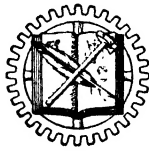# ENGINEERING

ʼHYSICAL METALLURGY

By

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SEVENTH PRINTING

Dedicated

TO MY WIFE
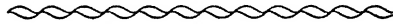
PREFACE

This volume is designed to aid those making their first acquaintance with engineering metals and alloys. Although published information of a high order is available, there have been few attempts to condense, correlate, and bring up to date the various materials, methods, and concepts of metals technology. The author has endeavored to produce a simple treatment of this field and to adapt it to the instruction of non-metallurgical students.

Although a wide range of materials within the major divisions of physical metallurgy has been included, no attempt has been made to approach the completeness of a handbook or to offer a laboratory manual. Selection of the subject matter and data was guided by the basic principles represented, as well as by their value as specific information.

The successful application of a new or improved alloy is not always a simple replacement operation. A modified or an entirely new design may be necessary to effect the greatest gain in utility, lowered cost, or improved appearance of the product. Furthermore, modification or complete change of the fabricating procedure may be required. In other cases it is necessary to adapt an alloy to the fabricating or service conditions at hand; hence alloys with " tailor-made " properties to meet the specifications of designers may be required. Under these conditions the need for cooperation of engineering and metallurgical personnel is quite evident.

In this book processing operations, including hot and cold reduction and fabrication, welding, heat treating, and machining, are discussed primarily from the metallurgical rather than from the operational viewpoint.

In view of the limited experience of many students in the mechanical testing of metals, the author first intended to add a supplement on the mechanical tests and their interpretation. However, as modern treatments of this subject are now available, this supplement has been omitted but the author strongly recommends

that tensile tests and the specialized tests for hardness, impact toughness, endurance (fatigue) limit, and creep strength be reviewed by the reader. Appropriate references on these points are cited in Chapter I.

In presenting data on mechanical properties, one is immediately confronted with the " yield point " problem. In this book the term " yield point " is applied to the stress at which a definite elongation is obtained without increase in load, and to all cases in which the original investigators reported a yield point even though it is quite certain from the nature of the material (its lack of a definite yield point) that the " dividers method " or some similar device had been used in the testing procedure. Although " yield strength " determinations based on a stated amount of deviation from the modulus line are highly preferred for such materials, and are now quite generally used, less recent data are, unfortunately, often given in terms of a less accurately defined " yield point " or " elastic limit."

The references to the technical literature include the sources of illustrations or other material and recent publications containing extended treatments of practically every topic discussed in this book. In nearly every case, the references are readily available publications in the English language. With certain exceptions, the most recent authoritative articles were selected rather than publications containing reports of original investigations.

The author is aware that certain faults and errors will be found in this book and he will be grateful to teachers who bring such errors to his attention.

May, 1939   R. H. H.

ACKNOWLEDGMENT

nellan of the Society for his kind cooperation. Special acknowledgment is also made the American Society for Testing Materials, the Society of Automotive Engineers, and the American Institute of Mining and Metallurgical Engineers for data from their publications.

The author is grateful to his wife for her painstaking work in typing the manuscript and checking the proofs.

ROBERT H. HEYER

MIDDLETOWN, OHIO,
May, 1939.

CONTENTS

CHAPTER I

PURE METALS

Many of our important metals including iron, copper, aluminum, zinc, lead, and nickel are used extensively in the pure or commercially pure state, and they will be discussed from this standpoint in subsequent chapters. In this chapter the pure metals will be used as examples in the study of the basic principles of atomic crystal structure, microscopic grain or crystal structure, plastic deformation, and recrystallization after cold working.

Crystal Structure of Copper. — Copper is a liquid at temperatures above 1083° C. Upon cooling it freezes at this temperature (1083° C.) to a crystalline solid. Under certain suitable conditions it will solidify as a single crystal in which the copper atoms are spaced according to the cubic pattern represented in Fig. 1.
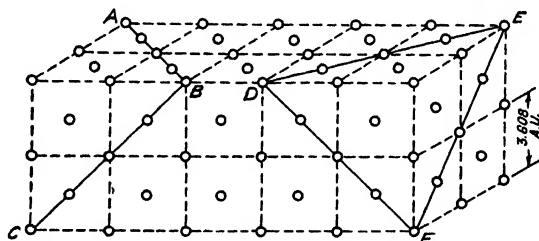
Fig. 1. — Arrangement of Copper Atoms in a Face-Centered Cubic Space Lattice.

This structure is known as the face-centered cubic (F.C.C.) space lattice, a type which is common to such well known metals as copper, aluminum, gold, nickel, and lead. The size of the cube is different for each metal. (See also Fig. 3, page 6.)

The spacing between the copper atoms along the edge of a unit cube has been accurately determined by X-ray diffraction methods [1] and has been found to be 3.608 A.U. when the metal is free from impurities and from internal stress. (A.U. = angstrom unit = 10^{-8} cm.) The relationship between the atomic structure and a

more familiar characteristic such as density is shown in the following calculation of the density of a perfect copper crystal.

A cube measuring exactly 3.608 A.U. on a side could be superimposed on a unit cell of the crystal structure so as to include one-half of each of the " face " atoms and one-eighth of each of the " corner " atoms of the cell. Although the complete cell is outlined by 14 atoms, the total number of atoms wholly within the superimposed cube is four. (See column 11, Table I.) The density of copper may be calculated from the number of atoms per cell, Avogadro's number, the volume of the cell, and the atomic weight of copper.

$$4 \, \frac{\text{atoms}}{\text{cell}} \times \frac{1}{6.06 \times 10^{23}} \frac{\text{mol}}{\text{atom}} \times \frac{1}{(3.608 \times 10^{-8})^3} \frac{\text{cells}}{\text{c.c.}} \times 63.57 \, \frac{\text{g.}}{\text{mol}} = 8.93 \, \frac{\text{g.}}{\text{c.c.}}$$

All actual metallic single crystals have been found to be somewhat imperfect. The deviation from true symmetrical construction within the crystal has been described as a mosaic of regularly occurring structural units which are slightly out of parallel. The cause and significance of this intracrystalline structure is still a matter of conjecture.[2, 3]

Plasticity of Crystals. — Experimentally prepared single crystals of plastic metals have certain notable properties. When bent or otherwise deformed they yield with extreme ease, block-like sections of a crystal gliding along adjacent sections, apparently tending to separate at the planes of movement. However, after severe deformation the crystal will remain intact and further distortion will cause slippage along new planes, indicating that the metal is actually stronger at the slip planes than in those parts of the crystal which have not yet yielded. The evidence is that after deformation the atoms along slip planes are quite disarranged as compared with those in undisturbed areas, and that upon standing at room temperature in this condition they may, in some cases, become more rigid and unyielding than they were just after deformation. Naturally these planes will have a general stiffening effect upon the entire structure, due, in part, to interference with further movement along potential slip planes whose paths cross the original planes of deformation. Strengthening by mechanical deformation is called " strain-hardening."

In face-centered cubic crystals it is possible for plastic deformation to occur along four families of planes of the type represented by ABC and DEF of Fig. 1. The possible directions of deformation in the plane DEF are represented by the three lines DE, EF, and FD. This is in conformity with a rule which is quite generally applicable, especially for face-centered cubic lattices, that slip occurs in the most densely populated plane and in the direction of the most
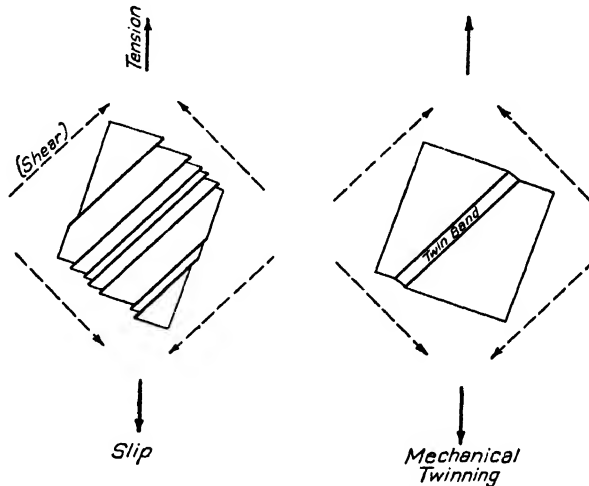
FIG. 2. — SINGLE CRYSTALS DEFORMED BY SLIP AND BY MECHANICAL TWINNING.

closely packed line within that plane. In any given case the external forces applied may be resolved into shearing stresses along the possible planes of slip and in the possible slip directions. The resulting deformation will occur first in the family of planes subjected to the highest shearing stress, provided this stress is greater than the critical value required to induce slippage. Considering a crystal deforming along a single plane at approximately 45 degrees to a single tensile force, it is apparent that this plane will soon be rotated with respect to the direction of the applied force and that other planes may become more favorably oriented with respect to the maximum shearing stresses.

The mechanism of plastic deformation is naturally more complex in ordinary commercial metals consisting of large numbers of crystals of varying orientations than it is in single crystals. In general,

TABLE I. PHYSICAL CONSTANTS OF IMPORTANT

1		2	3	4	5	6	7	8
Symbol	Element	Atomic Weight (1938)	Density g./cm.3 at 20° C.	Melting Point °C.	Melting Point °F.	Specific Heat-a	Linear Coef. of Thermal Expansion- b	Thermal Conductivity c
Li	Lithium	6.94	0.53	186	366.8	0.79	56	0.17
Be	Beryllium	9.02	1.85	1285	2345	0.425	12.3	0.3847
Mg	Magnesium	24.32	1.74	651	1204	0.249	25.7	0.37
Al	Aluminum	26.97	2.70	657	1214.6	0.2259	24	0.52
Pb	Lead	207.21	11.34	327.4	621.2	0.030	29.5	0.083
Sb	Antimony	121.76	6.62	630.5	1166.9	0.0493	11.29	0.0444
Bi	Bismuth	209.00	9.80	271	519.8	0.0290	13.45	0.0200
Sn	Tin	118.70	7.30	232	449.6	0.054	—	0.157
Cd	Cadmium	112.41	8.65	320.9	609.6	0.0547	29.8	0.217
Zn	Zinc	65.38	7.14	419.5	787	0.09	e	0.268
Hg	Mercury	200.61	13.55	−38.9	−38.0	0.0332	—	0.0200
Cu	Copper	63.57	8.94	1083	1981.4	0.0918	16.42	0.923
Fe	Iron, alpha	55.84	7.87	1535	2795	0.1075	11.9	0.19
Fe	Iron, gamma	55.84	—	—	—	—	—	—
C	Carbon	12.01	2.22	—	—	0.165	1.2	0.057
Si	Silicon	28.06	2.4	1427	2600	0.1762	—	0.20
P	Phosphorus	31.02	1.82	44.1	111.4	0.177	125	—
S	Sulphur	32.06	2.07	113	235.4	0.175	67.48	0.00063
Mn	Manganese	54.93	7.44	1242	2268	0.107	23	—
Ni	Nickel	58.69	8.9	1452	2646	0.112	13.7	0.140
Co	Cobalt	58.94	8.9	1490	2714	0.0989	12.08	0.165
Cr	Chromium	52.01	7.14	1550	2822	0.12	8.1	0.165
V	Vanadium	50.95	5.68	1710	3110	0.1153	—	—
Mo	Molybdenum	95.95	10.2	2620	4748	0.0647	5.49	0.350
W	Tungsten	183.92	19.3	3370	6098	0.034	4.0	0.476
Ti	Titanium	47.90	4.5	1800	3272	0.142	7.14	—
Zr	Zirconium	91.22	6.4	1700	3092	0.066	6.3	—
Cb	Columbium	92.91	8.57	1950	3542	—	7.2	—
Ta	Tantalum	180.88	16.6	2850	5162	0.0356	6.5	0.130
Ag	Silver	107.88	10.5	960.5	1761	0.0558	18.9	0.974
Au	Gold	197.2	19.3	1063	1945.4	0.0308	14.4	0.7072
Pt	Platinum	195.23	21.45	1773	3224	0.0319	8.8	0.166
Rh	Rhodium	102.91	12.44	1966	3571	0.0598	8.9	0.213
Pd	Palladium	106.7	12.0	1554	2831	0.0587	11.6	0.161
Ir	Iridium	193.1	22.4	2409	4368	0.0322	6.41	0.141

a Specific heat in cal./(g. ° C.) at room temperature. [Equals B.T.U./(lb. ° F.)
b Linear coefficient of thermal expansion per degree C. at room temperature —
c Thermal conductivity in cal/(cm. ° C. sec.) at room temperature.
d Not included in Metals Handbook.
e Varies from 23.0 across the grain to 32.5 with the grain for pure hot rolled zinc.

Courtesy of American Society for Metals. Condensed from Metals Handbook, p. 78

ALLOY FORMING ELEMENTS

9 Electrical Resistivity microhm — cm.	10 Type of Crystal Lattice at 20° C.	11 No. of Atoms in a Unit Cell-d	12 13 Lattice Constants in Angstrom Units at 20° C. (A.U. = 10⁻⁸ cm.)		14 Modulus of Elasticity lb./sq. in.
			" a "	" c "	
8.5	B.C.C.	2	3.51	—	—
18.5	Hex.C.P.	2	2.281	3.577	42 700 000
4.46	Hex.C.P	2	3.2022	5.1991	6 250 000
2.655	F.C.C.	4	4.0413	—	10 000 000
20.65	F.C.C.	4	4.9389	—	2 560 000
39	Rh.Hex.	2	4.4974	57°6'	11 300 000
115	Rh.Hex.	2	4.736	57°14'	4 600 000
11.5	B.C.Tetragonal	4	5.819	3.175	—
7.59	Hex.C.P.	2	2.9727	5.6061	—
6.0	Hex.C.P.	2	2.659	4.936	—
95.8	—	—	—	—	—
1.682	F.C.C.	4	3.6080	—	16 000 000
9.8	B.C.C.	2	2.8610	—	30 000 000
—	F.C.C.	4	3.60	—	—
1000	Hex. (Graphite)	4	2.46	6.69	—
85 × 10³	Cubic (Diamond)	8	5.4173	—	16 000 000
10¹⁷	—	—	—	—	—
1.9 × 10¹⁷	F.C.Orthorhombic	—	10.48	24.55	—
—	Cubic (Complex)	58	8.894	—	—
6.9	F.C.C.	4	3.517	—	30 000 000
9.7	Hex.C.P.	4	2.507	4.072	—
13.1	B.C.C.	2	2.878	—	—
26	B.C.C.	2	3.033	—	—
4.77	B.C.C.	2	3.1403	—	50 200 000
5.48	B.C.C.	2	3.1585	—	60 000 000
—	Hex.C.P.	2	2.953	4.729	—
41	Hex.C.P.	2	3.223	5.123	—
20.0	B.C.C.	2	3.2941	—	—
15.5	B.C.C.	2	3.2959	—	27 000 000
1.62	F.C.C.	4	4.0774	—	10 300 000
2.42	F.C.C.	4	4.0700	—	11 300 000
9.83	F.C.C.	4	3.9158	—	21 400 000
4.93	F.C.C.	4	3.7957	—	42 500 000
10	F.C.C.	4	3.8817	—	17 000 000
6.08	F.C.C.	4	3.8312	—	74 700 000

at room temperature.]
multiply all values by 10⁻⁶.

Modulus of elasticity and other properties also vary with direction.
(1939).

the most plastic metals and alloys are those whose crystals have face-centered cubic structures.

Although plastic deformation generally takes place by slip along various favorably located atomic planes, it may also be accomplished by a mechanism known as mechanical " twinning " [4], which may be considered to be a more definitely ordered type of deformation than simple slip. A schematic comparison of the two processes is given in Fig. 2. The relative displacements of the successive planes involved and certain other conditions necessary for mechanical

F.C.C.	B.C.C.	HEX.
Face-centered cubic	Body-centered cubic	Close packed hexagonal
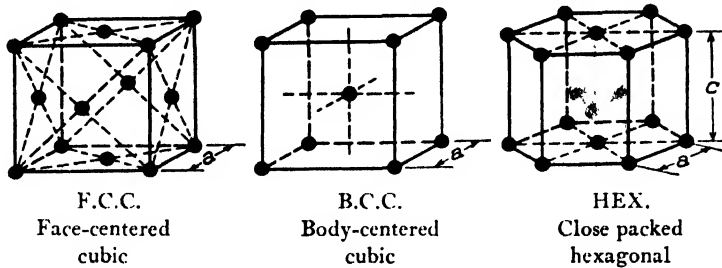
FIG. 3. — SPACE LATTICE TYPES.

twinning are definitely fixed for a given type of crystal structure. After deformation the atomic structure of the twinned portion (or portions) of the crystal is always identical with that of the original crystal and has a fixed crystallographic orientation with respect to the unchanged portion. The interface between twinned and original portions is unlike that between two separate crystals in that the atoms at twinning planes have definite and regularly spaced positions while those at grain boundaries are highly disordered. The twinning action is believed to occur simultaneously along all the planes involved in the formation of a single twin. When bars of tin are bent, the sudden internal displacements accompanying mechanical twin formation produce the cracking noise known as " tin cry." Sudden shocks sometimes produce mechanical twins in iron. (See page 153.)

From the standpoint of commercially important pure metals the three most common space lattice types are those shown in Fig. 3. The actual sizes of the unit cells, as they occur in the crystal structures of the pure metals, will be found in columns 12 and 13 of

Table I, page 5, in which " a " and " c " refer to the atomic spacings of Fig. 3.

The metals zinc and magnesium are important examples of the hexagonal space lattice. Single crystals of these metals are unusual in their plastic behavior because at room temperature they deform along but one family of planes — those parallel to the base plane

FIG. 4(a). — PLAN VIEW OF TRANSITION ZONE IN A SPECIAL ZINC TENSILE SPECI-MEN. X2.6. (*Courtesy of Richard F. Miller.*)
FIG. 4(b). — SIDE VIEW OF TRANSITION ZONE. X2.6.

of the hexagonal prism represented in Fig. 3. Polycrystalline metals and alloys having hexagonal structures, and other crystal structures with less symmetry, have relatively low ductility and may be quite brittle because of the small number of planes of easy slip within the individual crystals.

Miller [5] has reported the results of a tensile test made at 180° C. on a one-half inch diameter rod of zinc containing a large crystal whose basal plane was inclined at an angle of 56.5 degrees to the axis of the test specimen. An elongation of 180% was at-

tained in the reduced section through gliding action along this single family of planes. Unlike a normal tensile test, the original round section became elliptical as illustrated in Fig. 4. The surface traces of the slip planes show the manner in which the deformation occurred. The wholly unreduced section at the right consists of normal polycrystalline zinc, separated from the single crystal at the left by a grain boundary perpendicular to the longitudinal axis. The restraining action of this boundary on the material in the tapered section is remarkable evidence of the strengthening effect of grain boundaries in normal metals and alloys.

As previously stated, the application of stress is not always accompanied by plastic deformation of a purely translational type. In the comparatively simple case represented in Fig. 4, basal planes which were originally inclined 56.5 degrees from the axis were rotated to a final position 17 degrees from the axis in the reduced section. It is evident that specimens consisting entirely of small grains having different orientations are subject to complicated rotational as well as translational distortion because of interaction between the grains or crystals.

Iron and the group of elements headed by chromium in Table I are important examples of metals having the body-centered cubic space lattice. The excellent plasticity of iron at room temperature is attested by the performance of commercially pure iron sheets in difficult forming and drawing operations. Nevertheless, metals having the body-centered cubic crystal structure are generally less plastic than metals such as gold, copper, and lead. Although deformation by slip within the body-centered cubic lattice of iron tends to follow the general laws of plastic deformation, it is complicated by the fact that there are three different types of planes along which slip occurs, two of which are not planes of highest atomic population.[6] The traces of slip planes on polished surfaces (specimen stressed after polishing) are often wavy as shown in Fig. 5. The relatively straight, parallel lines of Fig. 6 are characteristic of slip planes in face-centered cubic metals. In each case the effectiveness of the grain boundary network in arresting or changing the direction of deformation is apparent.

Failure by Rupture. — Severe plastic deformation must ulti-

mately lead to failure by rupture even in the most ductile metals or single crystals. In less plastic metals failure is likely to take place by cleavage along definite crystallographic planes within the crystals, and brittle metals may fail in this manner before any appreciable permanent deformation is effected. In other cases separation occurs at the grain boundaries. Fatigue cracks generally traverse

Left. Fig. 5. — Traces of Slip Planes in a Polished Surface of Iron. (Etched with 3% nital before stressing.) X500. (*Courtesy of Charles R. Taylor.*)

Right. Fig. 6. — Traces of Slip Planes in a Polished Surface of Yellow Brass. (Unetched.) X100.

the grain structure without regard for grain boundaries. High temperature failures and fractures produced by impact are often intergranular. The temperature and the method of application of the load are important factors in determining the mechanism of failure.[7]

The Grain Structure of Metals. — In the preceding discussion it has been stated that commercial metals consist of large numbers of grains or crystals having varying orientations. The origin and nature of such structures will now be considered.

In the solidification of a pure metal from its molten state or phase,

crystallization ordinarily begins at a great many centers so that small grains or nuclei form at random orientations in the melt. As freezing progresses these nuclei grow in size through attachment of additional atoms which ultimately build up crystals having the same type of crystal structure and the same orientation as their nuclei. However, the external forms of these crystals will be influenced by mechanical interference from adjacent crystals and by certain directional growth characteristics to be considered later. Because of the differences in orientation from grain to grain there is no possibility of perfect matching and coalescence of these crytals as the last portions of molten metal freeze, hence films or grain boundaries of disarranged atoms must function as the bonding material. Even after drastic mechanical and heat treatments have been applied to the casting, we expect to find some evidence of grains, large or small, held together by this film of atoms which has been likened to an amorphous cement. In Fig. 6 the outline of the grains was made apparent by the presence of slip planes within the grains. A specimen of high purity lead has been polished and etched to show the grain boundaries in Fig. 7.[8] (This specimen was cold pressed and annealed after casting.)

Fig. 7. — Grain Structure of Lead. (Etched with 60% glycerine, 20% acetic acid, 20% nitric acid.) X100.

As demonstrated in Fig. 4, the grain boundary network may be expected to strengthen the metal because of its tendency to confine deformation within the individual grains. In general, the strengthening effect of grain boundaries originates from two sources. First, the boundary zone itself offers much greater resistance to slip than the regions within a crystal. Second, in passing across a boundary to a grain of different orientation, a potential slip must change direc-

tion, thus in fine grained metals the irregular path of plastic deformation introduces additional slip resistance. It is not surprising, therefore, that a normal polycrystalline metal is much stronger, in the usual sense, than a large single crystal, and that we associate fine grain size with high strength in our commercial metals. In fact, this association is so strong that we sometimes fail to recognize certain even more important factors which contribute to the strength of many of our alloys. Further reference to these strengthening factors will be made in later chapters.

Dendritic Solidification. — The mechanism of solidification of metals is somewhat more complicated than stated in the above section. After the establishment of solid crystal nuclei in the melt, the growth of these crystals is more rapid in certain fixed directions than in others. In cubic metals, for example, the directions of rapid growth are mutually at right angles in space, and in hexagonal metals six branches are often

Fig. 8. — Dendrites at the Surface of a Lead Casting. (Unetched.) X7.

formed, making 60 degree angles in a single plane. The angular formations which result are known as dendrites because they often resemble trees in appearance.

The interlocking dendrites of Fig. 8 were found in the shrinkage cavity of a lead casting. Relief structures of this type are often formed at free surfaces of castings because shrinkage during the last stages of solidification reveals the dendrites in skeleton form.

In castings of very pure metals it is often difficult to observe dendritic formations in polished and etched cross sections. In nearly all cast alloys and impure metals the original dendritic formations are preserved and intensified by chemical segregation. Copper castings, for example, may contain cuprous oxide particles unless the

metal is fully deoxidized before pouring; and because these particles freeze at a lower temperature than the primary copper dendrites, they are found in the interstices between the dendrites, as shown in Fig. 9. (See also page 110.) Several other examples of dendritic structures in castings will be found in subsequent chapters.

Recrystallization of Cold Worked Metals. — The elements copper and lead react in different ways to mechanical deformation at room temperature. A block of pure soft copper may easily be flattened by hammering or pressing. Indentation hardness tests will show that a considerable increase in hardness has resulted due to the strain-hardening action described above. When a similar block of lead is hammered at room temperature, no appreciable change in hardness will be found. Likewise deformation of the copper block carried out at 600° C. will result in comparatively little change in hardness.

FIG. 9. — COPPER CASTING CONTAINING APPROXIMATELY 0.1% OXYGEN. (Etched with H_2O_2 in NH_4OH.) X100.

Copper at 600° C. and lead at room temperature do not strain-harden permanently because they undergo spontaneous reformation or recrystallization of their grain structures after application of stress. In this process the formerly existing strain-hardened grain structure is transformed into a new system of unstrained grains having properties comparable to those of the original undeformed metal. Metals which are sufficiently plastic to withstand considerable deformation without rupture have a rather definite temperature range of recrystallization; for example, 200°–225° C. for pure copper and somewhat *below room temperature* for pure lead. To understand the mechanism of recrystallization (to be illustrated in more detail later in this section) it is necessary to consider the structural state

of the metal after cold work and the effect of heat on this structure. Severe cold deformation produces not only slip bands within a given crystal but a high degree of distortion amounting to reorientation of individual portions or fragments of the original crystal. Upon heating, the atomic mobility is increased and many of the differently oriented crystal fragments may act as nuclei for growth of new crystals, thus it is possible to form a large number of small recrystallized grains from a larger single crystal distorted by cold work.

A rational division point between hot and cold working operations may be made at the recrystallization temperature. Thus copper may be cold worked at room temperature and hot worked at 600° C. while lead undergoes hot work even at room temperature.

Sykes [9] has recently summarized the generally accepted variables influencing this process of recrystallization:

(1) " Recrystallization temperature is lower (a) the greater the amount of deformation (b) the smaller the grains prior to deformation (c) the purer the metal (d) the lower the temperature at which deformation is effected and (e) the longer the time of heating.

(2) " In any given metal the grain size after complete recrystallization is normally smaller the lower the recrystallization temperature.

(3) " Above the recrystallization temperature the grain size is normally greater the higher the temperature and the longer the exposure.

(4) " Certain conditions of nonuniformity sometimes give rise to the formation of abnormally large grains. This is called ' germination.'

(5) " Nonuniformities giving rise to germination are grain size contrast, strain gradients, temperature gradients, concentration gradients, and obstruction gradients.

(6) " Presence of foreign matter as a second phase introduces mechanical obstruction to grain growth. Resistance increases as the quantity of foreign matter or other phase increases and as the particle size decreases.

(7) " Presence of a certain amount of obstruction matter promotes germination.

(8) " The higher the germinative temperature the larger will be the grains and the quicker they will form.

(9) " The more rapid the heating through the germinative range, at least within limits, the finer will be the grain size."

Many of the implications of these laws must be omitted or only briefly discussed here. (A detailed treatment of this subject and the other principles introduced in this chapter will be found in " The Science of Metals " by Jeffries and Archer.[10]) Several of the more important factors will, however, be illustrated by tests made on cold worked iron.

In experiments by Kenyon and Burns[11] soft iron sheets were reduced in thickness by cold rolling various amounts, which resulted in an increase in tensile strength and decrease in ductility as indicated in the stress-strain curves of Fig. 10. The effect on the grain structure of similar material cold reduced 60% is shown in Fig. 11 in which individual

FIG. 10. — EFFECT OF COLD REDUCTION ON THE SHAPE OF THE STRESS-STRAIN CURVE OF ARMCO INGOT IRON SHEETS. (*Courtesy of Reid L. Kenyon and Robert S. Burns.*)

crystals have been elongated and reduced to small fragments not clearly visible when magnified 150 times; however, the general distortion within the grains is quite evident. Heating for four hours at 1000° F. (538° C.) resulted in partial recrystallization as shown by the appearance of small rounded grains in Fig. 12. Heating for 20 hours at 1000° F. (538° C.) produced further recrystallization, as shown in Fig. 13. An equal time at 1200° F. (649° C.) resulted in somewhat larger and completely recrystallized grains of more uniform size as shown in Fig. 14. It may be noted that the grain size obtained after this treatment is finer than that of the original material before cold rolling; however, this is not always the case since very fine grained metals may also be cold worked and recrystallized. The hardness values given with the illustrations show the marked softening effect which always accompanies recrystallization of cold worked metals.

Left. FIG. 11. — ARMCO INGOT IRON SHEET REDUCED 60% BY COLD ROLLING. (Etched with 3% nital.) Rockwell " B " 102. X150. (The samples used for Figs. 11 to 14 were processed by R. L. Solter.) *Right.* FIG. 12. — COLD REDUCED SHEET HEATED FOR FOUR HOURS AT 1000° F. AND COOLED AT THE RATE OF 25° F. PER HOUR. (Etched with 3% nital.) Rockwell " B " 74. X150.
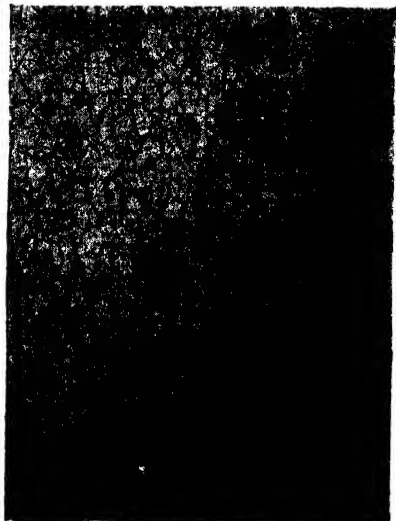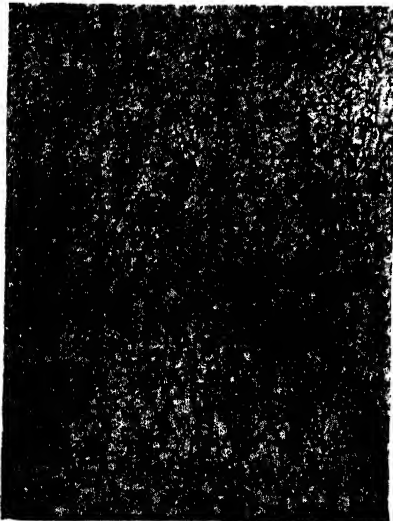
Left. FIG. 13. — COLD REDUCED SHEET HEATED 20 HOURS AT 1000° F. AND COOLED AT THE RATE OF 25° F. PER HOUR. (Etched with 3% nital.) Rockwell " B " 54. X150. *Right.* FIG. 14. — COLD REDUCED SHEET HEATED 20 HOURS AT 1200° F. AND COOLED AT THE RATE OF 25° F. PER HOUR. (Etched with 3% nital.) Rockwell " B " 41. X150.

Relatively fine grain sizes, as in Figs. 13 and 14, generally result upon recrystallization of severely cold worked metals. Much coarser structures (of lower hardness) may be obtained by establishing strain gradients within the metal before heating. (See item 5 above.) For example, a reduction of only 5% in thickness followed by heating to 800° C. gave a very coarse grain size in the experiments of Oberhoffer and Oertel (Fig. 15). This is an ex-
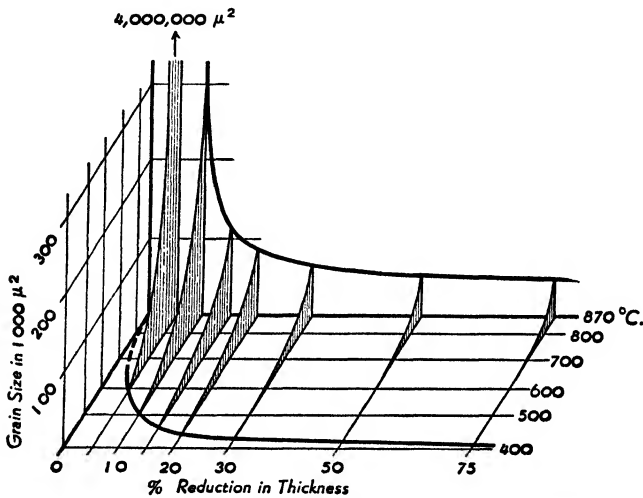
FIG. 15. — RECRYSTALLIZATION OF IRON. (*Experiments of Oberhoffer and Oertel reported by Kenyon and Mehl.*[12])

ample of one type of " germination." Other conditions which may lead to extremely coarse grain sizes are indicated in items 5, 6, 7, and 8 above.

Kaiser and Taylor [13] have recently shown that recrystallization characteristics of iron do not vary with the type of cold deformation (rolling, tension, or torsion) provided all deformations are expressed in terms of octahedral shear strains.

The general subject of recrystallization is important from the standpoint of annealing practice, especially for cold worked metals and alloys. For example, the development of extremely coarse grain size due to germinative conditions is generally undesirable and steps may be taken to eliminate or avoid these conditions, pos-

sibly by rapid heating to temperatures above the germinative temperature range.   (See item 9 above.)

The examples used to discuss solidification, plastic deformation, recrystallization, and grain structure have, in most cases, been pure metals.   Nevertheless, the principles involved apply to a great many alloys, including the important group known as solid solutions. Included among the well known alloys of this class are the yellow brass of Fig. 6, monel metal, austenitic stainless steel, and a large variety of copper base alloys in addition to ordinary brass.   (See Figs. 51 and 52 for quantitative data on the physical properties of cold worked and recrystallized brass.)   All of these materials will be found to have simple, homogeneous microstructures.   More complex alloys containing two or more microconstituents are, in general, less capable of plastic deformation and softening by recrystallization.

Cooling Rate after Recrystallization. — When a cold rolled sheet of pure iron or copper is heated above its recrystallization temperature, the newly formed grain structure will be retained without appreciable change regardless of the rate of cooling from the heat treating temperature.   There is no tendency to refine the grain structure or to harden the metal even at very high cooling rates such as produced by quenching in water.   In the case of copper and other nonferrous metals this result is seldom contested; however, the phenomenon of recrystallization is sometimes confused with the phase transformations which occur in hardening and annealing treatments used for steels and certain other alloys.   At this time it need only be noted that there is no change in the basic type of crystal structure during recrystallization, whereas the structure changes from the body-centered cubic to the face-centered cubic lattice upon heating steels through the critical transformation range for hardening purposes.   In the case of iron and steel the critical transformation temperatures (see page 276) are from 200° to 500° C. above the recrystallization temperatures; consequently, cold reduced iron and very low carbon steel sheets may be softened commercially by recrystallization at temperatures between 500° and 700° C. (932°–1292° F.) without heating through the critical transformation range.

Directional Properties Produced by Mechanical Deformation and Recrystallization. — When plastic metals are mechanically worked in such a way that all the deformation is in one direction, as in drawing wire, rolling bars, or rolling strip, the grains become elongated as well as fragmented and distorted, and the mechanical properties of the product assume directional differences. If the deformation is carried out above the recrystallization temperature range these effects are minimized and in some cases eliminated by spontaneous recrystallization or annealing of the structure during working. Cold deformation results in a fibrous structure, as in Fig. 11, in which the mechanical properties, especially ductility, are generally lower in the transverse than in the longitudinal direction.

FIG. 16. — FRACTURE OF A STEERING ARM SHOWING FAILURE BY FATIGUE. X2.

Directional properties produced in this way are caused not only by the obvious elongation and breakdown of the original crystals, but by arrangement of the crystal fragments in certain preferred orientations with respect to the boundaries of the piece and the direction of working. For example, one of the preferred orientations of crystal fragments produced during cold rolling of low carbon steel sheets may be described as follows, referring to Fig. 3(b): the top and bottom faces of the unit cell are parallel to the plane of the sheet and a diagonal of one of these faces is in the direction of rolling. In this material at least two other " preferred " positions are taken by the crystals [14], whereas in alpha brass, a typical F.C.C. alloy, the crystals tend towards a single orientation when severely deformed.[15] X-ray diffraction methods are the only means available for direct determination of the existence and nature of these deformation textures.

The extent to which textures form varies with the amount and type of deformation and with the metal being worked. In extreme

cases practically all of the deformed metal assumes a preferred orientation.

Recrystallization of severely deformed metals does not, as one might think, entirely eliminate the possibility of directional properties. Although the grains developed by recrystallization may be equi-axed or nondirectional in shape, they are likely to assume preferred orientations because their nuclei were ordered. The type of preferred orientation produced after recrystallization is not ordinarily the same as after deformation, as shown, for example, in the case of recrystallized cold rolled low carbon sheet steel by Gensamer and Lustman.[16]

It is interesting to note that in the case of alloys which develop marked deformation and recrystallization textures, the modulus of elasticity is subject to directional variations, despite the fact that the

Fig. 17. — Polished Section at the Fracture of an Overheated Forging. (Unetched.) Rockwell "C" 60. X100.

value of this constant undergoes little if any change with ordinary variations in analysis, heat treatment, or mechanical work.

Fatigue Failures. — Before leaving this general discussion of crystal structures and recrystallization it will be well to note that failures of machine parts by recrystallization in service at or near room temperature have never occurred, despite the fact that repeated statements to the contrary are on record, even in the technical press. All of the alloys, ferrous and nonferrous, used for working parts of machines are fully crystalline when put in service and are not subject to crystal growth under stress at normal temperatures.

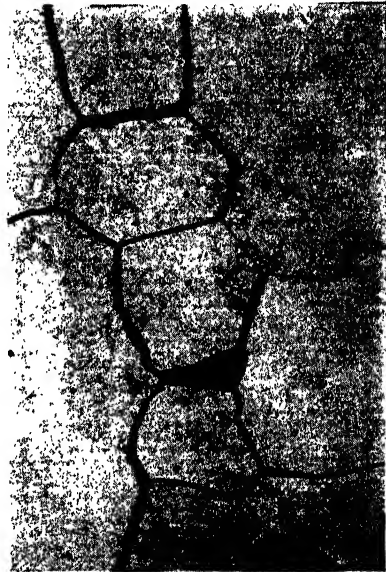An example of a fatigue failure of an automotive steering arm is

shown in Fig. 16. Cracking began at the regions of highest stress at the edges and gradually spread towards the center of the section. The fracture is comparatively smooth in the outer zones because these surfaces were rubbed together possibly thousands of times before sudden fracture of the central zone resulted in complete failure of the part. The relative coarseness of the suddenly fractured zone might be used as evidence of crystallization in service; however, it could easily be shown that a fresh fracture of the entire section would have the same appearance.

As one metallurgist of long-standing reputation has said, the crystallization failure experts must be allowed to die out gradually before the theory itself may be laid to rest.

A true example of a brittle failure due to coarse crystallization is shown in Fig. 17. This, however, is a case of overheated steel. The fracture was definitely intergranular in nature as shown by the cavities between the crystals.

Mechanical Properties — Metallographic Technique. — The general subject of mechanical properties and testing may be reviewed in recent books by Churchill [17] and Bullens. [18]

Metallographic laboratory methods are fully described by Kehl [19] and Vilella. [20]

REFERENCES

1. Clark, G. L., " Applied X-Rays," McGraw-Hill Book Company, New York (1932)
2. Desch, C. H., " The Chemistry of Solids," 77, Cornell University Press, Ithaca, N. Y. (1934)
3. Northcott, L., " Veining and Mosaic Structure in Metals," Metal Progress, **31,** 152 (1937)
4. Mathewson, C. H., " Twinning in Metals," Trans. Am. Inst. Mining and Met. Eng., **78,** 7 (1928)
5. Miller, R. F., " Influence of a Grain Boundary on the Deformation of a Single Crystal of Zinc," Trans. Am. Inst. Mining and Met. Eng., **111,** 135 (1934)
6. Barrett, C. S., Ansel, G., and Mehl, R. F., " Slip, Twinning and Cleavage in Iron and Silicon Ferrite," Trans. Am. Soc. for Metals, **25,** 702 (1937)
7. White, A. E., Clark, C. L., and Wilson, R. L., " The Fracture of Carbon Steels at Elevated Temperatures," Trans. Am. Soc. for Metals, **25,** 863 (1937)
8. Hodge, J. M., and Heyer, R. H., " The Lead-Zinc Eutectic," Metals and Alloys, **2,** 297 (1931)
9. Sykes, W. P., " New Light on Old Metallurgical Problems," Mining and Metallurgy, **20,** 54 (1939)

10. Jeffries, Z., and Archer, R. S., " The Science of Metals," McGraw-Hill Book Company, New York (1924)

11. Kenyon, R. L., and Burns, R. S., " Autographic Stress-Strain Curves of Deep-Drawing Sheets," Trans. Am. Soc. for Steel Treating, **21**, 577 (1933)

12. Kenyon, R. L., and Mehl, R. F., " Effect of Cold Work on the Properties of Iron," Metals Handbook, 453, Am. Soc. for Metals, Cleveland (1939)

13. Kaiser, H. F., and Taylor, H. F., " The Effect of Type of Cold Deformation on the Recrystallization Properties of Armco Iron," Trans. Am. Soc. for Metals, **27**, 227 (1939)

14. Gensamer, M., and Mehl, R. F., " Preferred Orientations Produced by Cold-rolling Low-carbon Sheet Steel," Trans. Am. Inst. Mining and Met. Eng., **120**, 277 (1936)

15. Pickus, M. R., and Mathewson, C. H., " Plastic Deformation and Subsequent Recrystallization of Single Crystals of Alpha Brass," Metals Technology, **5**, T. P. No. 989 (Dec. 1938)

16. Gensamer, M., and Lustman, B., " Preferred Orientations Produced by Recrystallizing Cold-rolled Low-carbon Sheet Steel," Trans. Am. Inst. Mining and Met. Eng., **125**, 501 (1937)

17. Churchill, H. D., " Physical Testing of Metals," Am. Soc. for Metals, Cleveland (1936)

18. Bullens, D. K., and Metallurgical Staff Battelle Memorial Institute, " Steel and Its Heat Treatment," Vol. I, John Wiley and Sons, New York (1938)

19. Kehl, G. L., " The Principles of Metallographic Laboratory Practice," McGraw-Hill Book Company, New York (1939)

20. Vilella, J. R., " Metallographic Technique for Steel," Am. Soc. for Metals, Cleveland (1938)

CHAPTER II

PRINCIPLES OF ALLOYING

It is possible for two pure metals to assume complicated structural relationships when they are mixed in the molten state and allowed to solidify. When three or more metals are involved, the structural complexity may become even greater. On the other hand, the structure of an alloy having three or four basic elements may, in some cases, be essentially homogeneous.

The present treatment of this subject will be simplified in that only binary or two metal systems will be included. The most important principles of alloying may be demonstrated by means of such systems, and many commercial alloys containing more than two elements may be effectively reduced to binary systems for structural considerations. In certain instances two metals form a complicated alloy system, but the alloys having complex structures lie outside the range of useful compositions.

Equilibrium Diagrams and Commercial Alloys. — For the study of alloy systems we have available a ready means of graphical representation known as the equilibrium or constitutional diagram. Such a diagram gives the temperature-composition relations of every alloy in a system, an invaluable aid in determining the possibilities of mechanical and heat treatment. Sometimes it appears as though these diagrams are rather limited by the requirement that the alloys be in a state of equilibrium. Commercial castings seldom attain equilibrium and, therefore, do not conform strictly to the diagrams. Nevertheless, we may make quantitative estimates and calculations from the equilibrium diagrams in order to establish certain limits which the alloy may approach under ideal conditions. Many mechanically worked and annealed products such as wire, sheet, forgings, pressings, and rolled shapes do reach structural equilibrium for all practical purposes and thereby conform satisfactorily to the diagrams. In the case of steel and many other alloys we may purposely and very forcibly upset the equilibrium to attain special results by heat treatment.

The Significance of Phases in Alloy Systems. — The term " phase," which has been used sparingly up to this point, is highly significant in the development and discussion of equilibrium diagrams. Findlay's[1] definition follows: ". . . a heterogeneous system is made up of different portions, each in itself homogeneous but marked off in space and separated from other portions by bounding surfaces. These homogeneous, physically distinct and mechanically separable portions are called *phases."*

According to Findlay's definition the phases are mechanically separable; however, in most alloys the experimental difficulties of separation would be great. For example, the mechanical separation of the cuprous oxide particles from the copper in the alloy of Fig. 9 would be a delicate procedure. At high magnification the oxide appears as small rounded particles embedded in copper. This does not mean that each particle is a separate phase. Copper and cuprous oxide are the phases present in Fig. 9.

Phase symbols for alloy systems are generally of the following types: A, B, AB$_4$, Cu, Al, CuAl$_2$, Fe$_3$C, α, β, δ, M$_A$, M$_{Fe}$, and M. Letters A, B, etc. are generalized symbols denoting pure metals in the solid state and are equivalent to the chemical symbols Cu, Al, etc. Pure metals in the molten state are represented by M$_A$, M$_{Fe}$, etc., and molten solutions of metals by the letter M. The formulas AB$_4$, CuAl$_2$, Fe$_3$C, etc. represent intermetallic compounds. The Greek letters α, β, etc. represent solid solutions of metals.

The significance of phases will become more apparent after their identification in a number of equilibrium diagrams and photomicrographs.

A Classification of Binary Metallic Alloy Systems. — There are several bases upon which binary systems may be classified. In this text the nature of the solid alloys formed will be taken as the primary factor in establishing a relationship between the elementary types of alloy systems. At one extreme are binary systems in which the two metals are insoluble after solidification, hence the resulting alloys are merely mechanical mixtures. At the other extreme are the systems in which the two metals are completely soluble after solidification, forming solid solution alloys. In either case the solubility relations in the liquid state may vary as follows:

Solubility in Solid State	Solubility in Liquid State	Example — Figure No.
Insoluble (Mechanical mixture alloys)	Insoluble	18
	Partial	(not treated in this chapter)
	Complete	19, 20
Complete (Solid solution alloys)	Insoluble	(highly improbable type)
	Partial	(highly improbable type)
	Complete	24

After consideration of the above types the more complicated systems involving partial solid solubility will be studied (Figs. 25 and 26), followed by those systems in which intermetallic compounds are formed (Figs. 27, 28, and 29).

A BINARY SYSTEM IN WHICH THE ALLOYS ARE MECHANICAL MIXTURES IN BOTH THE LIQUID AND THE SOLID STATES

Upon cooling a crucible of molten metal A from 800° C. the temperature will vary with cooling time as in Fig. 18(a). The interval of time at constant temperature is caused by evolution of heat of crystallization at the freezing point. In the same figure may be found a similar freezing curve for pure metal B. The remaining curves in Fig. 18(a) suggest that each of these metals has
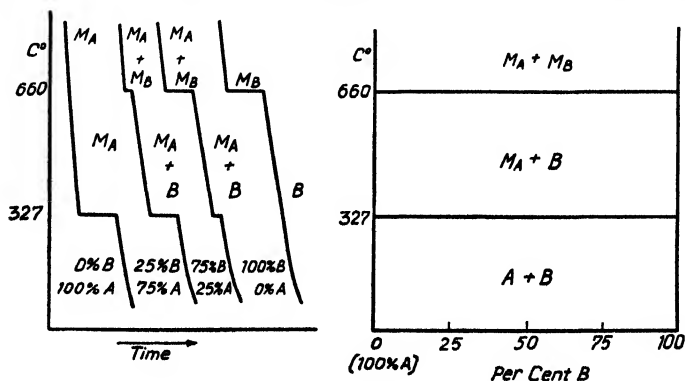
Left. FIG. 18(a).—COOLING CURVES FOR ALLOYS OF A AND B.
Right. FIG. 18(b).—AN EQUILIBRIUM DIAGRAM FOR METALS A AND B WHICH FORM MECHANICAL MIXTURES IN BOTH THE LIQUID AND SOLID STATES.

no effect on the freezing temperature of the other. The shortened constant temperature lines simply indicate smaller amounts of A or B present in the alloys, assuming that the total charge in the crucible remains constant in all cases.

If we transfer the information of Fig. 18 (a) to the temperature-composition coordinates of Fig. 18 (b), we have a more workable diagram. Metals A and B are insoluble liquids at 800° C.; a combination of molten A and solid B at 500° C.; and a simple mechanical mixture of solid A and B at 300° C. and lower temperatures. Assuming that A is much denser than B, a slowly cooled alloy would consist of two distinct layers with pure A at the bottom of the crucible and pure B at the top.

It is sometimes stated that simple mechanical mixtures of this type are thermodynamically impossible and that all properly melted alloys have some solubility in the solid state. Admittedly this solubility may be extremely slight and incapable of representation in diagrams of ordinary scale. Thus we may justify many of the " thermodynamically impossible " diagrams which follow on the basis of practical expediency. The possibility of complete insolubility in the liquid state is even more remote. The diagram of Fig. 18, for example, is a fairly close representation of the lead-aluminum system (where Pb = A and Al = B), however it has been shown that lead is soluble in liquid aluminum to the extent of about 1.5 % at 660° C., the melting point of aluminum, and about 13.0 % at 1000° C.[2] This complicates the equilibrium diagram for our present purpose, hence hypothetical metals having zero liquid solubilities have been used for illustration.

A BINARY SYSTEM IN WHICH THE ALLOYS ARE MECHANICAL MIXTURES IN THE SOLID STATE AND SOLUTIONS IN THE LIQUID STATE

The metals bismuth and copper come very close to forming an alloy system such as that drawn in Fig. 19 (b) (where Bi = A and Cu = B). However, the generalized symbols A and B will again be used. The diagram as drawn represents the simplest case in which two metals are completely soluble in the liquid state and insoluble in the solid state.

The symbol M may now be used to denote any molten solution

from 0 to 100% B. The curved line joining the melting points of pure A and pure B indicates the temperature at which solid B crystals begin to form during freezing of any alloy in the system. Apparently A has a definite tendency to lower the freezing temperatures of its alloys with B. Examination of the cooling curves of Fig. 19(a) reveals the fact that B crystallizes over a wide range of temperatures (due to the influence of A dissolved in the melt)
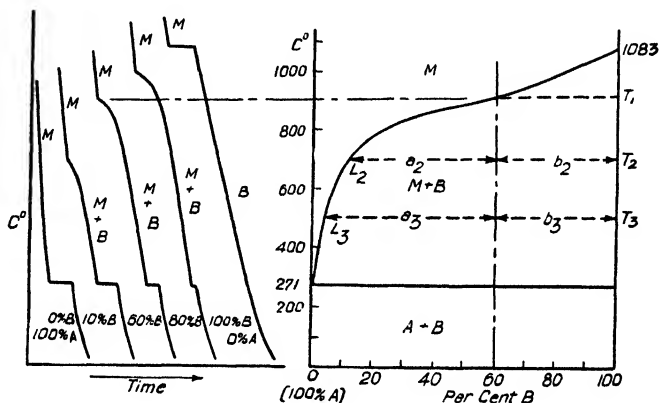
Left. FIG. 19(a). — COOLING CURVES FOR THE ALLOYS OF A AND B.

Right. FIG. 19(b). — AN EQUILIBRIUM DIAGRAM FOR METALS A AND B WHICH FORM MECHANICAL MIXTURES IN THE SOLID STATE AND SOLUTIONS IN THE LIQUID STATE.

rather than at any fixed temperature as in the system illustrated in Fig. 18.

The mechanism of solidification will be followed, using the 60% B — 40% A alloy as an example. When temperature T_1 is reached on cooling, pure solid B begins to freeze from the melt. However, as soon as an appreciable amount of B freezes, the composition of the remaining molten metal changes towards a lower B content. When the alloy cools to temperature T_2 the molten portion loses enough B through crystallization to reach the composition L_2 (approximately 10% B — 90% A). Upon further cooling the chemical composition of the molten portion changes to L_3 at temperature T_3 and finally to pure A at 271° C.

Determination of the Chemical Compositions of the Phases from the Equilibrium Diagram. — In any single phase region the com-

position of the phase is the same as the composition of the alloy.

Two phase regions are bounded on the right and left by single phase regions. In general, the *chemical composition* of one of the phases present in a two phase region may be found by noting the intersection of a horizontal line, at the selected temperature, with one of the boundary lines of the two phase region. In Fig. 19(b), for example, the composition of the molten phase at temperature T_2 is represented by the intersection of the horizontal line with the boundary of the molten region at L_2. Obviously the composition of the pure metal B, at the opposite intersection, is 100% B. Note that at constant temperature the chemical composition of the alloy as a whole may vary considerably, within the boundaries of the two phase region, without affecting the *chemical compositions* of its two phases. On the other hand, the *proportions* of the phases present will depend upon the exact chemical composition of the entire alloy.

Determination of Proportions of Phases Present in Two Phase Alloys. — Upon cooling the 60% B alloy to temperature T_2, the molten portion attains a composition indicated by L_2, as in the above example. By using the relative lengths a_2 and b_2 we may calculate the proportions of solid and molten phases present. The principle of the calculation is not difficult. At temperature T_2 the relative amount of B varies from 0% at composition L_2 to 100% at the boundary line at the right, and the ratio $\dfrac{a_2}{a_2 + b_2}$ expresses the proportion of B present in the alloy considered; therefore

$$\frac{100\,a_2}{a_2 + b_2} = \frac{100 \times 50}{50 + 40} = 55.5\% \text{ pure solid B, and}$$

$$\frac{100\,b_2}{a_2 + b_2} = \frac{100 \times 40}{50 + 40} = 44.5\% \text{ molten metal}$$

Similar calculations at temperature T_3 show that an increased proportion of the alloy has solidified as pure B by the time this temperature is reached.

At 271+° C., just above the horizontal line at 271° C., the state of the 60% B alloy may be summarized as follows. Solid B, which

has crystallized between temperatures T_1 and 271° C., is mixed with molten metal which has now become practically pure A. The relative proportions of these phases are:

$$\frac{100 \times 40}{60 + 40} = 40\% \text{ molten A}$$

$$\frac{100 \times 60}{60 + 40} = 60\% \text{ solid B}$$

Upon cooling from 271+° C. to 270° C., all of the molten A solidifies at the constant temperature 271° C., the normal freezing point for metal A. The significance of the cooling curves in Fig. 19(a) and the phase symbols of Fig. 19(b) should now be more readily apparent.

The method of determining the *proportions of the phases present,* as demonstrated above, may be applied to any region of any binary equilibrium diagram providing two phases are present in that region. One must make sure that the ratios selected are in the right order. If the single phase region of that phase whose proportion is to be calculated lies to the left of the two phase working region, use the right hand or " b " distance in the numerator, and vice versa. A few trial calculations will show that this inverse relationship is necessary.

A simple device for solving such problems may be prepared by the reader. A white elastic tape or rubber band may be stretched to two or three times its normal length, and while thus extended a uniform scale from 0 to 100 by tens may be inked on the tape. Two or three basic sizes may be used so that there are available scales of infinite variability from about ¾ inch to 5 inches over-all length. By placing the zero line of such a scale at L_3, for example, and the 100 line at T_3, the per cent B may be read directly from the scale at its intersection with the original composition line; 60% B — 40% A.

Segregation in Mechanical Mixture Alloys. — The mechanical mixture alloys of the systems illustrated in Figs. 18 and 19 are subject to extreme segregation provided there is a marked difference in the densities of the metals involved. If metals A and B of Fig. 19 have nearly the same density, we should expect a fairly uniform distribution of the solid A and B crystals, especially if solidification

is rapid. For example, in an alloy which contains only a small amount of A we might find the crystals of solid A present in the form of a film or coating joining the primary B crystals which were the first to freeze on cooling. If pure B is naturally very plastic but A is hard and brittle, such a film or network would have a very detrimental effect on the plasticity of the alloy. This happens to be the condition which is responsible for the extremely poor wire drawing characteristics of copper containing very small amounts of bismuth.
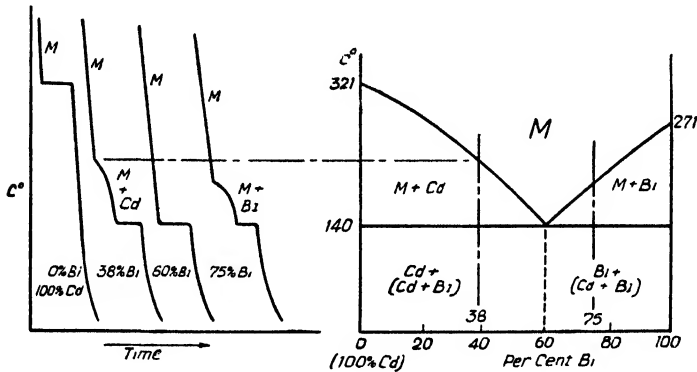
Left. FIG. 20(a). — COOLING CURVES FOR BISMUTH-CADMIUM ALLOYS.
Right. FIG. 20(b). — THE BISMUTH-CADMIUM EQUILIBRIUM DIAGRAM.

MECHANICAL MIXTURES OF THE EUTECTIC TYPE

Usually when two metals are soluble in the liquid state each metal very definitely influences the freezing of the other, resulting in the formation of the familiar eutectic and other more or less complicated types of binary systems.

An important class of alloy systems is represented by the Bi-Cd diagram of Fig. 20(b). Bismuth lowers the temperature at which cadmium begins to crystallize, and cadmium has the same effect on bismuth. The alloy whose composition is indicated by the intersection of the two curved boundary lines is called the eutectic alloy, and the temperature of final solidification is called the eutectic temperature (140° C.).

The eutectic temperature line is also known as the solidus in this case since only solid phases appear below it. The curved line is
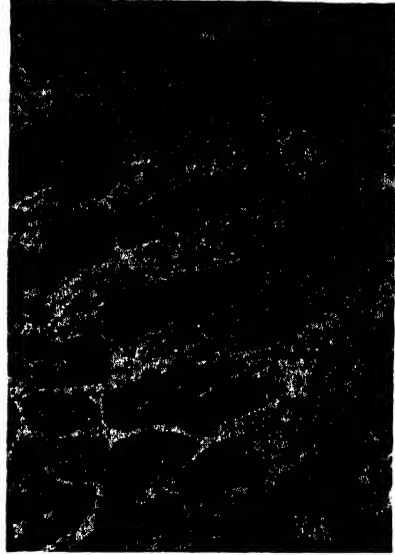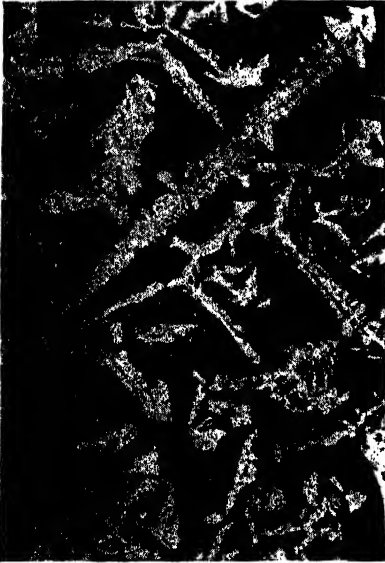
Left. FIG. 21 — A CAST ALLOY CONTAINING 75% BI — 25% CD. X100. (Figs. 21 to 23 etched with ferric chloride — HCl.)

Right. FIG. 22. — A CAST ALLOY CONTAINING 38% BI — 62% CD. X200.
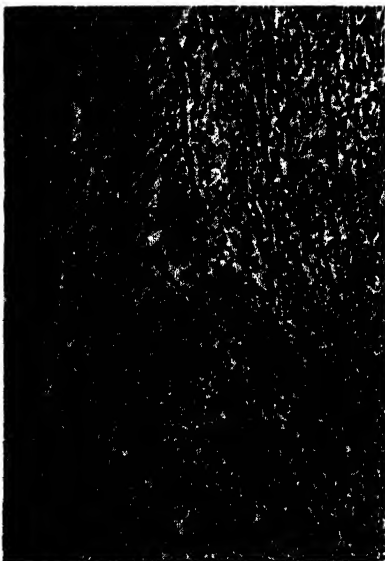
FIG. 23. — A CAST ALLOY CONTAINING 60% BI — 40% CD. X400.

called the liquidus since only the liquid phase exists above it (neglecting the vapor phase which is always present, and, at sufficiently high temperatures, is the only phase present).

The 75% Bi alloy may be cooled to the liquidus line, whereupon pure bismuth crystals will begin to freeze. The crystallization of bismuth is then continuous down to the eutectic temperature. Here we find that

$$\frac{100(75-60)}{100-60} = 37.5\%$$ of the

alloy is pure solid bismuth, the remainder being molten metal whose chemical composition is 60% Bi —

40% Cd. All of this molten metal finally solidifies at the eutectic temperature, producing the horizontal line in the fourth cooling curve of Fig. 20(a).

The nature of the resulting alloy is apparent in the photomicrograph, Fig. 21, in which the eutectic portion appears as a rather finely divided mixture of cadmium and bismuth crystals. The large white portions of this structure are *primary* bismuth crystals, which solidified between the liquidus and the solidus as rather well defined dendritic units. Very much smaller bismuth crystals are also present in the eutectic. The distinction between the primary and the eutectic bismuth crystals is physical rather than chemical. Cadmium appears only in the eutectic portion of this alloy. Cadmium and bismuth may be distinguished in the photomicrograph because the etching reagent used, acid ferric chloride, has a preferential or selective chemical action on the cadmium, coloring it dark. Nearly all etching reagents operate on the principle of selective attack.

The 38% Bi — 62% Cd alloy of Fig. 22 is structurally similar to the 75% Bi alloy. The primary cadmium phase indicated on the equilibrium diagram is the dark etching portion of the microstructure. It should be noted that the chemical composition of the speckled eutectic portion is the same as in Fig. 21. The coarser appearance of the eutectic in Fig. 22 is due to higher magnification of the photomicrograph and a lower cooling rate in casting.

The 60% Bi alloy freezes completely at the eutectic temperature, giving a single horizontal line in the cooling curve. Its finely divided structure is shown in Fig. 23. Eutectics in metal systems appear in several characteristic forms in which the phases are intimately mixed and usually finely divided. The eutectic structure of Fig. 23 consists of a mechanical mixture of two phases present in the following proportions at room temperature:

$$\frac{100\ a}{a+b} = \frac{100 \times 60}{60 + 40} = 60\%\ \text{Bi phase}\ (40\%\ \text{Cd phase})$$

All eutectic mixtures will be indicated by means of brackets around the phase symbols as in Fig. 20(b).

Definition of Components and Constituents. — The components of an alloy system are the pure metals of which it is composed.

The term " constituent," as used in this book, is described by Archer as follows: " Those portions of an alloy which under the microscope appear to be definite units in the structure are called constituents." [a]   In Figs. 21 to 23 the constituents are primary cadmium, primary bismuth, and eutectic, while the components are the metals cadmium and bismuth.

Structural Analysis in Terms of Constituents. — The relative amounts of the phases present in two phase regions of Figs. 19 and 20 have been calculated using a simple proportionality ratio.   The same procedure may be used to determine the relative amounts of the *constituents* present in any binary alloy at any temperature. Analysis by structure rather than by phases present is often of great practical value.   For example, in the laboratory it is frequently necessary to estimate the chemical composition of an alloy by its appearance in the microscope.   With no contributory information available, such as a qualitative chemical analysis or a knowledge of the source of the alloy, this is often a hopeless task.   In certain types of alloys nothing short of quantitative chemical analysis will give the answer.   An estimate is possible, however, providing the following requirements are met.

(1)   The qualitative chemical analysis must be known.

(2)   The equilibrium diagram must be known.

(3)   The alloy must be in equilibrium.

(4)   The alloy must be in a two phase region of the diagram at the given temperature.

(5)   The constituents (pure metals, eutectics, eutectoids, solid solutions, intermetallic compounds) must be distinguishable in the microscope in order to estimate their relative proportions.

Fortunately all of these conditions are frequently fulfilled and it is then a matter of seconds to make a reasonable estimate of the chemical composition.   As an example, any annealed plain carbon steel may be readily classified as to carbon content by observation of its microstructure.

Example of Estimation of Chemical Composition from the Microstructure. — By visual estimation, Fig. 22 appears to contain approximately 60% eutectic and 40% primary cadmium in its micro-

structure. If it were 100% primary cadmium it would, of course, analyze 0% bismuth. On the other hand, if it were 100% eutectic in structure it would analyze 60% bismuth. These figures represent the compositions of the constituents, primary cadmium and eutectic, and may be taken as the working limits. The composition of the actual alloy will fall somewhere between these limits. If the reader has an elastic measuring tape at hand he may place the zero mark at the 0% bismuth line and the 100 mark on the 60% bismuth line. The estimated 60% eutectic structure may then be located on the tape and the corresponding alloy, 36% bismuth, noted on the equilibrium diagram. By calculation the same result is obtained from the relationships:

$$a + b = 60\% \; ; \; \frac{a}{a+b} = 0.60 \text{ eutectic}$$
$$a = 60\% \times 0.60 = 36\% \text{ bismuth}$$

The estimate of the proportions of the constituents present was reasonably correct because the alloy contains 38% bismuth by chemical analysis.

Another source of error, in addition to estimating the proportions of constituents present, is the unequal densities of the structures estimated. The densities of bismuth and cadmium are 9.75 and 8.64 respectively, thus the density of the eutectic will not be exactly equal to that of pure cadmium. Since the estimates from the photomicrograph were actually made on the basis of relative areas, whereas the diagrams are drawn on a relative weight basis, density should be taken into account in more accurate estimates. (When the constituents are well defined in the photomicrograph, it is possible to cut out one of them with a scissors, using a large print if possible, and to compare the weight cut out with the total weight of the print.)

Solid Solutions. — Metals are not always insoluble in the solid state. All of the alloys in certain binary systems are completely soluble in both the liquid and the solid states. Included in this classification are the binary systems Cu-Ni, Au-Ag, Ni-Co, and Mo-W. Before studying alloy systems of this type the nature of solid solution will be considered.

We may think of a simple solid solution as a very intimate but statistically random association of two or more types of atoms in the solid state, the possible number of solute atoms present being restricted, in many cases, by solubility limits. The associations between the solute and solvent *atoms* are within the crystal lattice of the solvent phase, a situation quite different from the mechanical association of distinctly different *crystals,* however small, in the mechanical mixture alloys.

As an example, let 100 tungsten atoms be arranged according to the body-centered cubic pattern of Fig. 3. Replace 10 tungsten atoms, at random, with 10 molybdenum atoms. The result approximates the construction of the individual crystals of an alloy containing 90% tungsten-10% molybdenum. The arrangement described is close to the ideal structure of the 90 atomic per cent tungsten-10 atomic per cent molybdenum alloy in which 10 atoms rather than 10 grams out of 100 are molybdenum. For practical reasons we do not ordinarily use the atomic per cent basis in equilibrium diagrams.

One further modification must be made to improve the accuracy of our sketch of the above alloy. Since the original spacing of the tungsten atoms (as the pure metal) was 3.1585 A.U., and the spacing of the molybdenum atoms was 3.1403 A.U., the crystal lattice of the tungsten will be somewhat contracted by the presence of 10% molybdenum in solid solution.

The above example is known as a substitutional solid solution of the normal or disordered type. In another substitutional type the solute atoms do not assume statistically random replacement positions, but take special positions in the solvent lattice so as to superimpose a new order of symmetry on that of the parent lattice, resulting in a structure known as a " superlattice." [4] In a third type of solid solution the solute atoms do not replace atoms of the parent lattice, but find new positions within the lattice. This type is called the interstitial solid solution.

It has been observed that the fundamental physical properties such as crystal structure, atomic weight, atomic volume, density, and melting point must not be widely divergent if two metals are to form a complete series of solid solution alloys. Our ability to predict

the type of alloys to be expected from the physical constants of the alloying elements is limited at present. Progress in this direction has been made, especially in the interesting field of intermetallic compound formation. Reports of work in the field of atomic physics indicate that valuable assistance may be expected in the future in the establishment of the equilibrium relationships in our alloy systems. X-ray crystal structure analysis is recognized as the outstanding contribution of the science of physics to the study of alloys. Mathematical physics may become a still more powerful tool of analysis. Any assistance in dealing with three, four, or five metal systems will be especially welcomed by metallurgists.
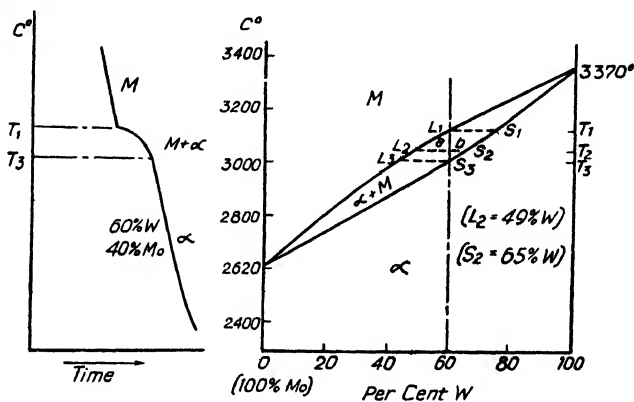
Left. Fig. 24(a). — Cooling Curve for the 60% W — 40% Mo Alloy.
Right. Fig. 24(b). — The Molybdenum-Tungsten Equilibrium Diagram.
(*International Critical Tables, Vol. II, p. 438.*)

A BINARY SYSTEM IN WHICH THE ALLOYS FORM SOLUTIONS IN BOTH THE LIQUID AND SOLID STATES

The molybdenum-tungsten alloys of Fig. 24(b) will be used to illustrate several of the features of solid solution systems.

A study of the solidification of the 60% tungsten alloy will aid in establishing one of the most important principles of alloying, that of *dendritic segregation.* Upon cooling to the liquidus of Fig. 24(b), a solid solution of composition S_1 will begin to freeze from the melt. This solid solution is richer in tungsten than the alloy as a whole, consequently we should look for a reduction in the percentage of tungsten in the remaining molten metal. In fact, as cooling pro-

gresses, the compositions of the melts are indicated by the intersections of the temperature lines with the liquidus, from L_1 to L_3, exactly as in the case of a mixture-forming alloy of the bismuth-cadmium type. The great difference in the mechanism of solidification of the present alloy and the mechanical mixture types of Figs. 18, 19, and 20 is the variability of the composition of the solid phase during freezing of the solid solution. According to the fundamental method for determining the chemical composition of a phase by noting the intersection of the temperature line with the appropriate boundary line, we find that the solid metal changes in composition along the solidus from S_1 to S_3. At temperature T_3 the alloy will have solidified and will be composed entirely of an alpha solid solution of composition S_3. Furthermore, the cooling curve of Fig. 24(a) contains no interval of constant temperature solidification, the gradual crystallization of the alloy between temperatures T_1 and T_3 being responsible for a gradual or progressive time delay in that portion of the curve.

The ideal solidification mechanism described above is seldom realized in commercial castings of solid solution type alloys. Solidification under equilibrium conditions requires a much lower cooling rate than is ordinarily possible. Time must be allowed for equalization of the chemical composition of all the solid metal present, and for establishment of equilibrium between this solid metal and the remaining melt. As the amount of the solid phase increases this state becomes more and more difficult to attain because of the constantly changing chemical composition of the successive layers of the solid portion, and the difficulty of obtaining rapid atomic diffusion throughout these layers.

Returning to the 60% tungsten alloy of Fig. 24, assume that cooling has been rapid down to temperature T_2. The solid alpha phase will vary in chemical composition from approximately 75 to 65% tungsten. Upon holding at this temperature it is possible for tungsten and molybdenum atoms to diffuse countercurrently until the composition becomes uniform throughout all of the solid phase present. The composition of alpha will then approximate the average value of 70% tungsten. This is not yet the equilibrium condition for this alloy at this temperature. In order to attain structural

equilibrium the melt and alpha phases must react until they reach the compositions L_2 and S_2 respectively. The melt will then contain 49% tungsten-51% molybdenum, and the alpha phase will contain 65% tungsten-35% molybdenum. Furthermore, the *proportions* of melt and alpha will be:

$$\frac{100\,a}{a + b} \cdot \frac{100\,(60 - 49)}{(60 - 49) + (65 - 60)} = 68.7\% \text{ alpha } (31.3\% \text{ melt})$$

It is possible to check these calculations as follows:

Phase Proportions	Phase Compositions		Per Cent of Each Element in the Alloy	
			W	Mo
68.7% alpha	{	65% W	44.7	
		35% Mo		24.0
31.3% melt	{	49% W	15.3	
		51% Mo		16.0
			$\overline{60.0}$ +	$\overline{40.0}$ = 100%

By cooling the alloy *very slowly* until entirely solidified it should be possible to obtain a perfectly homogeneous, single phase, alpha solid solution alloy of the composition, 60% tungsten-40% molybdenum. Microscopically it would be indistinguishable from a pure metal. Faster cooling prevents homogenization by diffusion and gives cored dendritic structures such as are often found in commercial castings. The equilibrium diagram should not be used quantitatively in such cases.

Microscopic evidence of dendritic segregation and the practical methods for homogenizing such structures will be demonstrated with commercially important metals in the chapter on copper base alloys.

A SOLID SOLUTION SYSTEM WITH A MINIMUM IN THE SOLIDUS CURVE

It is possible to form a complete series of solid solutions in which the alloys have melting points lower than those of the two pure metals. The copper-gold alloys, of considerable interest in dentistry, belong to this class.[5]

All of the solid solution systems mentioned up to this point involve pairs of metals having the same type of crystal structure, either B.C.C. or F.C.C.   When metals such as chromium, B.C.C., and nickel, F.C.C., form solid solutions, a minimum in the liquidus and solidus curves is likely, and it is questionable whether the solutions form a continuous series, since at some composition, or composition range, there must be a transition from the one basic type of crystal lattice to the other, and an intermediate region of two phase alloys is inevitable.
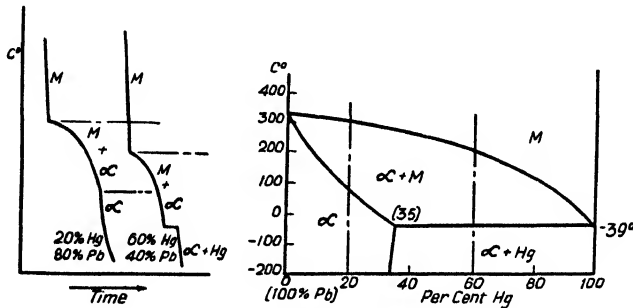
Left. FIG. 25(a). — COOLING CURVES FOR LEAD-MERCURY ALLOYS.
Right. FIG. 25(b). — THE LEAD-MERCURY EQUILIBRIUM DIAGRAM.   (*International Critical Tables, Vol. II, p. 414.*)

Steel at High Temperatures as a Solid Solution Alloy. — It is intended that many of the principles developed here shall be applied later to irons and steels.   For example, steel at heat treating temperatures is a solid solution of carbon and other elements in F.C.C. iron.   The carbon atoms in this case do not replace iron atoms in the crystal lattice but find special locations in the interstices between the iron atoms.   (See page 34.)   Interstitial solid solutions of this type allow freer movement of the solute atoms within the crystals than is generally possible in the substitutional types.

A SYSTEM IN WHICH THE SOLID SOLUBILITY IS LIMITED

In Fig. 25, which has much in common with Fig. 19, the factor of solid solubility has been introduced in a mixture system.   Mercury is soluble in lead up to 35% at —39° C., the freezing point of mer-

cury. When alloys with greater mercury contents are cooled below
—39° C. they consist of saturated alpha solid solution mixed with
crystals of solid mercury. The solidification of alloys containing
less than 35% mercury is directly comparable to that of the tung-
sten-molybdenum type solid solution alloys, hence dendritic segre-
gation may be expected if cooling is rapid. The cooling curve for
the 60% mercury alloy, see Fig. 25(a), shows the usual time delay
during the separation of primary alpha solid solution between the
liquidus and the solidus, followed by the constant temperature line
for solidification of pure mercury crystals at —39° C. The primary
alpha solid solution constituent of this alloy is also subject to den-
dritic segregation.

Under equilibrium conditions the proportion of mercury crystals
present in the 60% mercury alloy at —40° C. is approximately:

$$\frac{100(60-35)}{100-35} = 38.5\% \text{ mercury}$$

The balance of the solid alloy is saturated alpha solid solution whose
chemical composition is 35% mercury, balance lead.
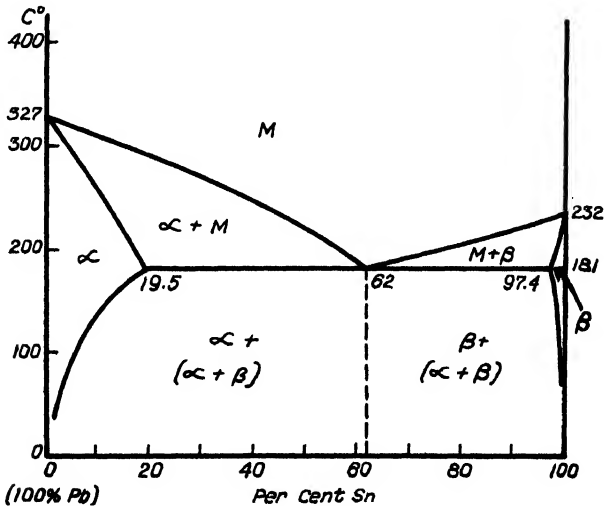
FIG. 26. — THE LEAD-TIN EQUILIBRIUM DIAGRAM. (*Metals Handbook*, 1939, *p.
1525.*)

The lead-tin diagram of Fig. 26 is representative of a large number of alloy systems characterized by mutual partial solid solubility. Upon cooling the 62% tin alloy to 183° C., a eutectic forms which consists of a mechanical mixture of two saturated solid solutions having the compositions: $\alpha = 80.5\%$ lead, 19.5% tin; $\beta = 2.6\%$ lead, 97.4% tin. As the eutectic alloy cools to room temperature, the solubility limits of both solid solutions decrease with temperature, and the compositions of the alpha and beta phases approach the pure metals lead and tin respectively.

The condition of mutual partial solubility is often difficult to visualize. No alloy in such a system ever contains crystals of the pure metals, although, in some cases, very small amounts of solute atoms are sufficient to saturate either solid solution.

It is suggested that the reader calculate the relative proportions of saturated alpha and beta phases present in the eutectic alloy just after its solidification (180° C.). An elastic scale may be applied between the limits 19.5 and 97.4% tin to obtain the proportion of beta directly.

Solid solution alloys having marked changes in solid solubility with temperature are subject to improvement in strength and hardness by the precipitation hardening method of heat treatment which will be described in a later chapter.

Soft solders are essentially lead-tin alloys in the range 15% to 63% tin, balance lead. The familiar 50% lead-50% tin composition and others are used to the extent of about 10,000 tons per year for joints in tin cans, automobile radiators, lead and other non-ferrous piping, etc. Gonser and Heath [6] and Hiers [7] give properties and other data for soft solders.

INTERMETALLIC COMPOUNDS IN BINARY ALLOY SYSTEMS

When two metals have a relatively strong affinity they are likely to form solid solutions or intermetallic compounds, or both. The compounds probably require a higher state of attraction than solid solutions. The elements present in intermetallic compounds unite

in definite atomic proportions, as in the case of true chemical compounds, but unlike the chemical compounds they follow no simple valence rules since all the metals have positive valence according to the usual conventions. The relationship between intermetallic and other compounds is discussed by Desch.[8]

Compounds which involve metals and metalloids such as Cu_3P and $FeSi$, as well as ordinary chemical compounds such as Cu_2O and FeS, may be treated in the same way as intermetallic compounds in the equilibrium diagrams.
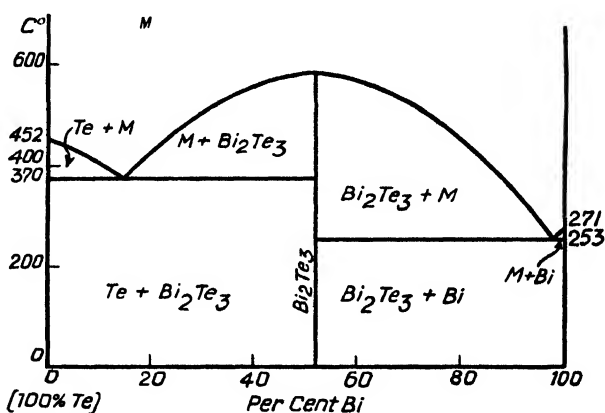
FIG. 27. — THE TELLURIUM-BISMUTH EQUILIBRIUM DIAGRAM. (*International Critical Tables, Vol. II, p. 428.*)

Although solid solutions may contain solvent and solute atoms in infinitely variable proportions, within the limits of solubility, and in a more or less random arrangement in the space lattice, intermetallic compounds form crystals in which the relative numbers of each type of atom and their relative positions in the lattice are fixed.

In general, the crystal structures of intermetallic compounds are more complex and have less symmetry than their components. This condition is responsible for the hardness and brittleness which is associated with intermetallic compounds. Among the important compounds which are responsible for the strength, hardness, and wear resistance of many of our most useful alloys are Fe_3C, $CuSn$, $CuAl_2$, WC, and Fe_4N. When intermetallic compounds are present in the form of particles of proper size, properly distributed in a
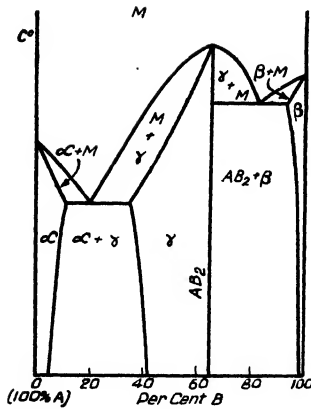
ductile matrix, the resulting alloys develop desirable mechanical properties not attainable in pure metals or solid solutions.

An intermetallic compound appears on the equilibrium diagram as a vertical line, and may be treated in the same manner as a pure metal in its structural relationships. In the microstructure of an alloy an intermetallic compound may appear as a separate constituent, or as a part of a complex constituent such as a eutectic mixture.

The equilibrium diagram of Fig. 27 will be recognized as a combination of two eutectic systems in which the intermetallic compound Bi_2Te_3 forms eutectic mixtures with each of the two pure metals. It will be noted that the symbols for the phases present, rather than the constituents, have been used in labeling the two phase regions below the eutectic temperatures.

The special features of Fig. 28 include partial solid solubilities of: AB_2 in A (α), AB_2 in B (β), and A in the compound AB_2 (γ). The resulting solid solutions form eutectic mixtures.

No solid solutions are shown in the Mg-Ni system of Fig. 29 which contains a compound, Mg_2Ni, which decomposes at 770° C., forming $MgNi_2$ plus M. The transformation which takes place at 770° C. in this diagram is known as a peritectic reaction, a type which is fairly common in alloy systems. At point P the following reaction takes place upon cooling:

$$MgNi_2 + M \rightarrow Mg_2Ni.$$

FIG. 29. — THE MAGNESIUM-NICKEL EQUILIBRIUM DIAGRAM. (*International Critical Tables, Vol. II, p. 437.*)

The same reaction applies to all alloys between A and B; however, between A and P an excess of melt remains and between P and B an excess of $MgNi_2$ remains after reaction is complete at 770° C.

The possible alloy systems have by no means been exhausted in this chapter. Notable omissions have been certain cases of *partial liquid solubility* (mentioned in discussing Fig. 18) and of phase changes in the solid state, the most important example of which, the *eutectoid transformation,* will be studied extensively in the iron-iron carbide system in Chapter VII.

REFERENCES

1. Findlay, A., " The Phase Rule and Its Applications," Longmans, Green and Co., New York (1931)
2. Kempf, L. W., and Van Horn, K. R., " Solubility of Lead and Bismuth in Liquid Aluminum and Aluminum-copper Alloys," Metals Technology, **8**, T. P. No. 990 (Dec. 1938)
3. Archer, R. S., " Constitution of Iron-Carbon Alloys," Metals Handbook, 366, Am. Soc. for Metals, Cleveland (1939)
4. Barrett, C. S., " A Correlated Abstract on Superlattices," Metals and Alloys, **8**, 251 (1937)
5. Doan, G. E., " Principles of Physical Metallurgy," McGraw-Hill Book Company, New York (1935)
6. Gonser, B. W., and Heath, C. M., " Physical Properties of Soft Solders and the Strength of Soldered Joints," Trans. Am. Inst. Mining and Met. Eng., **122**, 349 (1936)
7. Hiers, G. O., " Soft Solders and Their Application," Metals and Alloys, **2**, 257 (1931)
8. Desch, C. H., " Bimetallic Compounds and ' Order-Disorder,' " Metal Progress, **33**, 479 (1938)

CHAPTER III

THE WHITE METALS AND THEIR BEARING ALLOYS

Most of the elements which are of basic importance as industrial metals or in compounding alloys for engineering purposes are included in the following groups: the white metals (second group in Table I, page 4); the light metals (first group in Table I); copper; iron; and the elements which may be present in irons and steels as alloy additions or impurities (carbon through columbium in Table I).

In addition to its use as an alloying element in steels, nickel is also used as a pure metal and as a base for alloy formation. Likewise chromium (as an electrodeposited coating), tungsten, and molybdenum have certain engineering applications in which they are not alloyed with iron. Silver, in the noble metal group of Table I, has recently been used for airplane engine bearings, and numerous other elements not included in the table have specialized applications.

The following chapters conform to the above classification in that each is devoted mainly to one of the basic groups, with the exception of Chapter V in which both the light metal and the white metal die casting alloys are reviewed. Chapters VII to XV deal with iron and the ferrous alloys.

The White Metals. — The principal white metals of commercial importance are included in the second group of Table I. Innumerable alloys including bearing metals, solders, type metals, die casting alloys, storage battery alloys, low-melting alloys, and pewter consist of various combinations of these heavy low-melting, white-colored metals. Tin, zinc, and cadmium coatings for steel are additional uses of the white metals to be considered in a later chapter. Die castings are also treated separately, leaving bearing metals as the principal topic for this chapter.

"Modern Uses of Nonferrous Metals"[1] is recommended as

a descriptive treatment of nonferrous alloys. (Chapters 4, 9, 19, and 21 are of immediate interest.) The entire field of bearing materials is covered in a book by Bassett.[2]

The relative costs of the principal white metals may be obtained from the following figures for January 4, 1939 (base cost in cents per pound) : lead 4.70, zinc 4.50, cadmium 60.0, tin 46.40, antimony 14.00, bismuth 105.0, and mercury 101.3. Although it is not in the white metal group, copper at 11.12 cents per pound may also be mentioned because it is an important element in babbitt.

The crystal structures of these elements are diverse as may be seen in Table I. The alloys formed tend towards mechanical mixture type structures and intermetallic compounds, with rather limited solid solubilities.

Structural Nature of Bearing Metals. — A large group of successful nonferrous bearing alloys is characterized by structures containing a hard, wear resisting phase or phases, embedded in a plastic, load-sustaining matrix or groundmass. The wear resisting qualities of the white metal bearing alloys are attributed to intermetallic compounds such as $CuSn$, $SbSn$, Pb_3Ca, and $NiCd_7$. The plastic matrix is usually a solid solution of Pb, Sn, or Cd base, or a binary or ternary eutectic in which one of these elements is the major constituent.

Another type of structure which has proved successful consists of a more or less homogeneous matrix which is sufficiently strong and wear resisting to meet the service requirements, and in which flakes, particles, or a network of a softer phase exists. Cast iron and copper-lead alloys, to be discussed in later chapters, are examples of this type of bearing alloy. In both types the heterogeneous structures provide surface irregularities of microscopic dimensions which, according to one school of thought, aid in the distribution of oil. On the other hand, it is said that a properly worn-in bearing develops a perfectly smooth surface because of its plasticity. In at least one instance, that of engine bearings lined with pure silver, a homogeneous and initially smooth surface is provided.

Requirements of White Metal Bearing Alloys. — The following list of requirements is evidence of the versatility expected of white metal bearing alloys.

(1) Wear resistance.

(2) Plasticity — to provide for slight irregularities in alignment and machining, and for deflections under service stresses.

(3) Compressive strength — to prevent squeezing or extrusion of the metal from the bearing box or liner. The strength and wear resistance which may be developed in a given alloy are limited by the plasticity requirements.

(4) Shock resistance — to resist cracking under unusual operating conditions and impact loads.

(5) Low frictional characteristics in contact with shaft material — to prevent galling or seizure and excessive wear under starting conditions, during periods of overloading, and in cases of partial failure of the lubrication system. This property is closely associated with " oiliness," or the capacity for retaining the oil on the bearing surface and preventing metal to metal contact.

(6) Capacity to embed abrasive particles which may pass between the bearing metal and the harder journal — to minimize the possibility of scoring the journal.

(7) Retention of strength and wear resistance at elevated temperatures — to provide for both normal and unusually high operating temperatures.

(8) High thermal conductivity — to reduce operating temperatures.

(9) Good bonding characteristics — to aid in heat conduction and to prevent mechanical separation of the bearing alloy from its container.

(10) Ease of remelting and recasting — an important requirement in railroad and other types of bearing service.

(11) Chemical stability — to prevent corrosion in air or by the lubricants used. This is often a matter of proper selection of the lubricant.

(12) Low cost.

Lead-Antimony Alloys. — The lead-antimony equilibrium diagram, Fig. 30, is of interest from the standpoint of several important lead alloys in addition to bearing metals. For example, tele-

TABLE II.   COMPOSITION AND PHYSICAL PROPERTIES OF WHITE METAL BEARING ALLOYS

A.S.T.M. Alloy No.	Specified Composition, per cent				Specific Gravity	Yield Point* lb./sq. in.		Brinell Hardness**		Melting Point °F.	Temperature of Complete Liquefaction °F.	Proper Pouring Temperature °F.
	Sn	Cu	Sb	Pb		68° F.	212° F.	68° F.	212° F.			
1	91.0	4.5	4.5	—	7.34	4400	2650	17.0	8.0	433	—	825
2	89.0	3.5	7.5	—	7.39	6100	3000	24.5	12.0	466	669	795
3	83.3	8.3	8.3	—	7.46	6600	3150	27.0	14.5	464	792	915
4	75.0	3.0	12.0	10.0	7.52	5550	2150	24.5	12.0	363	583	710
5	65.0	2.0	15.0	18.0	7.75	5050	2150	22.5	10.0	358	565	690
6	20.0	1.5	15.0	63.5	9.33	3800	2050	21.0	10.5	358	531	655
7	10.0	—	15.0	75.0	9.73	3550	1600	22.5	10.5	464	514	640
8	5.0	—	15.0	80.0	10.04	3400	1750	20.0	9.5	459	522	645
9	5.0	—	10.0	85.0	10.24	3400	1550	19.0	8.5	459	493	620
10	2.0	—	15.0	83.0	10.07	3350	1850	17.5	9.0	468	507	630
11	—	—	15.0	85.0	10.28	3050	1400	15.0	7.0	471	504	630
12	—	—	10.0	90.0	10.67	2800	1250	14.5	6.5	473	498	625

* From compression tests made on cylinders 0.5 inch diameter by 1.5 inches machined from chill castings 0.75 inch diameter by 2.0 inches. Values taken from stress-strain curves at a deformation of 0.125% reduction of gage length.

** Brinell tests made on the bottom surface of parallel machined specimens cast in a 2 inch diameter by 0.625 inch deep steel mold at room temperature. A 500 kg. load was applied to a 10 mm. ball for 30 seconds.

Condensed from an appendix to Standard Specifications B23–26, A.S.T.M. Standards, Part I, p. 704 (1936). Additional data are given in the original table. The tests were made at the National Bureau of Standards. The compositions of the actual alloys tested (as given in the original table) were close to the specified compositions, the maximum deviation for any one element being 1.0%.

NOTE: The data in the above table do not constitute a part of Specification B23–26. They are merely given to indicate the physical properties which can be expected of carefully manufactured alloys of the formulas indicated.

phone cable sheath is generally a 1% antimony-lead alloy. This composition falls within the alpha region of limited solid solubility and is, therefore, subject to age-hardening or may be hardened by heat treatment if desired. (See page 69.) The alloys containing 6 to 12% antimony, sometimes known as " hard leads," are used extensively for storage battery grids, and in the form of sheet and extruded pipe they have many applications of a chemical nature."

The microstructures of the alloys containing less than 12½% antimony consist of primary lead-rich solid solution crystals (alpha) and the harder lead-rich eutectic ($\alpha + \beta$).
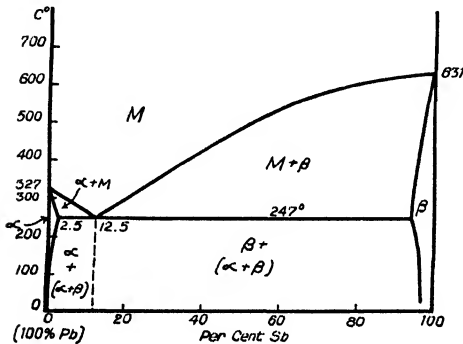
FIG. 30. — THE LEAD-ANTIMONY EQUILIBRIUM DIAGRAM. (*Metals Handbook, 1939, p. 1516.*)

Lead-Antimony Bearing Alloys. — The cheapest bearing metals are lead-antimony alloys modified by tin and in some cases by copper. Very small amounts of arsenic and iron may be present. Typical compositions of lead base bearing alloys, as well as tin base types to be considered later, are given in Table II. These alloys are made in a wide range of compositions including many not given in the table.

The microstructure of an alloy whose composition approximates A.S.T.M. No. 10 is shown in Fig. 31. It consists almost entirely of the eutectic constituent. The structure of a commercial alloy having a very high antimony content, Fig. 32, contains a large proportion of white, angular crystals. In Fig. 31 the white crystals are probably antimony containing both lead and tin in solid solution; however, the compound SbSn, which has the same appearance, may also be present. In Fig. 32 it is quite likely that the primary crystals are essentially the intermetallic compound SbSn. In either case the primary crystals are known to be hard, wear resisting constituents, and although there may be uncertainty as to their composition, the microscope is useful in checking their size and distribution. In bearing metals of this general type, including tin base alloys, it is

known that a certain optimum range of particle size gives best bearing characteristics for a given type of service. Extremely large units may score the journal, and, due to the necessarily small number of such particles present in a given composition, the matrix will be left in a relatively soft and weak condition. On the other hand, extremely fine particle size may so stiffen the otherwise plastic matrix that it cannot carry out its normal functions.
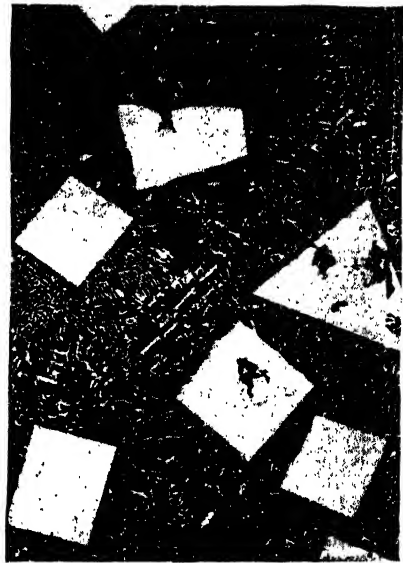
Left. Fig. 31. — LEAD BASE BEARING METAL CONTAINING 15% SB AND 3% SN. Brinell 18. X100.

Right. Fig. 32. — LEAD BASE BEARING METAL CONTAINING 29% SB AND 12% SN. Brinell 24. X100.

The lead base bearing metals are widely used for railway applications, heavy machinery, pump shafts, line shafts, and other purposes where bearing stresses and speeds are moderate, and size and weight requirements are of minor importance. The lead base alloys are more susceptible to failure by softening and melting than the other types to be discussed. The low melting point of the lead-antimony eutectic is largely responsible for this characteristic.

Genuine Babbitts — Tin Base Bearing Alloys. — The genuine babbitts are *tin base* alloys. Their excellent reputation is reflected in their continued wide usage in competition with several newly

developed alloys of high merit such as the cadmium base types. They are commonly used at bearing loads as high as 15–1800 lb. per sq. in. and in some cases up to 2500 lb. per sq. in. They have definite advantages over all lead base types at high peripheral speeds and under severe impact conditions.

The trend in bearings for gasoline engines is towards the use of thin steel shells containing a layer of bearing alloy which is sometimes as thin as 0.010 inch. The processes of cleaning and tinning the strip steel, applying the babbitt to the strip, cutting to length, and forming and machining the half bearings are illustrated by Fisher in Metal Progress.[4]

Cast bearings with thicker sections may be poured into a steel or cast iron journal box, properly prepared by tinning, and frequently preheated to about 200° F. A mandrel may be used to complete the mold. Mold and pouring temperatures are important in controlling the size and distribution of the hardening particles.

In recent years babbitt bearings have also been made by centrifugal casting of very thin layers of the white metal on bronze or steel bearing shells. The commonly accepted advantages of centrifugal casting; namely, controlled cooling, uniformity, and freedom from porosity, are thus secured. Spraying methods have been developed whereby a thin coating of bearing alloy is applied to a specially prepared surface by melting a babbitt wire in an oxyacetylene spray gun.[5]

The compositions of genuine babbitts are more nearly standardized than those of the lead base types. The A.S.T.M. alloys 1, 2, and 3 of Table II, page 47 are typical. The compositions having lower copper and antimony contents have the higher fluidity, which is an important requirement in casting bearings with very thin sections. Higher copper and antimony contents give greater hardness and strength. Typical applications and mechanical properties of babbitts, as well as the effects of impurities such as zinc, arsenic, aluminum, and bismuth on the properties of these alloys have been summarized by Ellis.[6, 7]

The microstructure of a tin base babbitt is shown in Fig. 33. According to a simplified interpretation of such structures the white cubical particles are the intermetallic compound SbSn, the white

needle-like phase is CuSn, and the matrix is essentially a solid solution of antimony and a small amount of copper in tin.[7]   The compound CuSn has a hexagonal crystal structure and frequently tends to solidify in the form of six-pointed stars.   It is the first constituent to freeze on cooling, and, because of its uniform distribution throughout the melt, it tends to prevent segregation of the SbSn cubes which freeze at a lower temperature.

FIG. 33. — TIN BASE OR GENUINE BABBITT CONTAINING 9% SB AND 7% CU.   Brinell 36.   X100.

If the observer is sure that lead and other extraneous elements are not present in appreciable amount and that the casting conditions have been uniform for the alloys examined, he can distinguish high or low antimony contents in genuine babbitts by the relative quantity of the SbSn cubes present. Likewise relative copper contents may be estimated by the presence of more or less of the needle-like constituent.   For a given composition the sizes of these hardening particles will vary greatly with the casting conditions.   The alloy whose structure is shown in Fig. 33 was cast as a ¼ inch section in a warm metal mold, which resulted in a rather coarse structure typical of heavy bearings.   Thin sectioned bearings of the type used for gasoline engines are cast under conditions which induce faster cooling and much smaller crystal sizes.

Lead-Calcium Bearing Alloys. — The shortage of tin and antimony in the World War was responsible for the development in Germany, and in this country, of a special series of lead base bearing metals.   These alloys are variously known as " alkali-alkaline earth lead-base," " precipitation hardening lead-base," and " lead-calcium " alloys.[8]   Typical compositions of two lead-calcium bearing alloys which are well known in the railway industry follow.

	Pb	Ca	Na	Li	K	Sn	Al
Bahnmetall	98.6	0.7	0.6	0.04	0.03	—	0.2 max.
Satco	97.4	0.15	—	—	0.07	2.4	—
Satco	97.5	*	—	—	—	1.0	—

* Hack describes Satco metal as a 1% tin-lead alloy, hardened with considerable calcium.[9]

The hardening agents in these bearing metals are intermetallic compounds such as Pb_3Ca, Na_2Pb_5, and possibly a hard Na–Ca compound in the Bahnmetall type. The hardness obtained is believed to be due largely to the extremely fine particle size of these compounds. (The details of hardening by precipitation of fine particles will be discussed in Chapter IV in connection with the heat treatment of the aluminum alloys.)

The lead-calcium alloys have, in several instances, given service equal to the tin base types, and they are definitely superior to the ordinary lead base types in load carrying capacity and resistance to pounding. They are more expensive than the cost of the component metals might indicate because of difficulty in introducing such elements as calcium and sodium into the melt. They are also very difficult to remelt without loss of the hardening elements. Furthermore, they are the least stable chemically in air and in the presence of corrosive lubricants. On the other hand, the elements which harden these alloys do not lower the freezing ranges as do antimony and tin in the ordinary lead base type. (Equilibrium diagrams are given by Grant.*) This is responsible for one important feature of the alkali-lead alloys; namely, their ability to withstand overheating to higher temperatures than either of the other types under consideration.

The principal application of the lead-calcium bearing metals has been in the railway field. Large quantities of Bahnmetall have been used in the German systems, and the Satco type has been used rather extensively in this country, both in steam and Diesel service.

Lead alloys hardened by calcium have been developed for use outside the field of bearing metals. The effect of small amounts of calcium in lead alloys is quite remarkable. Alloys containing only 0.1% calcium are sufficiently strong for use in storage battery plates and telephone cable sheaths.[10]

Significance of Brinell Hardness of Bearing Metals. — The

hardness and frictional qualities of the alkali-lead alloys are equivalent to those of genuine babbitts. However, indentation hardness is not recognized as a faithful measure of the wear resistance of white metal bearings. The 10 mm. ball of the Brinell test forces the microscopic wear resisting particles through the plastic matrix irrespective of the hardness of the particles themselves. Obviously the result is more closely related to the load carrying capacity of the bearing than to its wear resistance. Most of the tin base and alkali-lead bearing alloys have hardnesses between 24 and 36 Brinell. The ordinary lead base types are softer, approximately 14 to 22 Brinell.

A scratch-hardness tester known as the Bierbaum microcharacter has been found useful in determining the relative hardnesses of the various microconstituents in bearing alloys.[11] A specially ground diamond point is drawn over a polished surface of the metal under a known load and the width of the resulting scratch is measured by examination at high magnification. Upon cutting across the hard wear resisting constituents the scratch becomes very much narrower than in the matrix. Relative scratch hardnesses of the microconstituents may thus be obtained, using a conversion formula developed for the test.

Cadmium Base Bearing Alloys. — The comparatively recent development of the cadmium base bearing alloys for use in internal combustion engines has been of considerable interest. The tin base babbitts, which have been standard for many years for engine bearings, first yielded ground to the copper-lead bearing metals, to be discussed later under copper alloys, and later to the cadmium base alloys. According to Smart[12] the principal disadvantage of the tin base babbitts is their susceptibility to fatigue failure at elevated operating temperatures. The higher operating speeds, loads, and temperatures of present day gasoline engines greatly increase bearing requirements, which led to the development of the new alloys. It is still too early to judge which type, if any, will eventually dominate the highly specialized automotive engine bearing field.

The cadmium base bearings withstand higher operating pressures than tin base babbitts, yet they do not tend to score the journal, which is a serious limitation of many harder materials for bearing service. The superiority of the cadmium base bearing alloys in

retention of strength and hardness at elevated temperatures has been associated with the relatively high melting temperatures of the cadmium-nickel and cadmium-copper eutectics compared with the tin-rich and lead-rich eutectics of other white metal bearing alloys. (The equilibrium diagrams for the cadmium alloys have been assembled by Smart.[12])

The mechanical performance of the cadmium base bearings has been excellent. Difficulties have, however, been reported which have been caused by corrosion in lubricating oils which contain or develop organic compounds of an acid nature. Corrosion of this type has occurred in copper-lead and certain lead base alloys as well as in the cadmium base. Means have been developed for inhibiting the formation of these compounds in lubricating oils. An interesting application of a rarely used element, indium, to overcome this corrosive action has been reported by Smart.[13] Indium is plated on the surface of the finish-machined bearing and then diffused into the alloy by heating to a low temperature. The amount used is less than 0.5% of the weight of the bearing metal.

The continued use of cadmium base alloys will depend upon their superior service characteristics, because their cost is greater than that of the genuine babbitts. The rather limited supply of cadmium, produced as a by-product of zinc smelting, was formerly used for electroplating and a few minor alloys. The cost of cadmium advanced from 55 cents to over one dollar per pound in 1937 and 1938 because of the demand for cadmium as a bearing metal. Increased capacity for the recovery of by-product cadmium has again made this metal available at a lower price.

The principal compositions developed and used in this country are listed below.

Identification		Per Cent Composition					Reference
		Ag	Cu	Ni	In	Cd	
Federal-	CS50	0.75	0.5	—	—	Balance	[14]
Mogul	CS51	2.5	0.3	—	—	Balance	[14]
Corp.	CS53	2.9	0.6	—	—	Balance	[14]
American Smelting		—	—	1.3	—	Balance	[15]
and Refining Company		—	—	3.0	—	Balance	[15]
General Motors Corp.		2.25	0.25	—	—	Balance	[12]
(Pontiac Motor Division)		2.25	0.25	—	0.2	Balance	[13]

The silver-copper types with 0.75 to 2.9% silver have been used extensively in this country. Silver is said to improve the casting characteristics by reducing oxidation and increasing fluidity. Structurally it forms a solid solution with cadmium, thereby increasing its strength but leaving it comparatively plastic. Copper is only slightly soluble in these alloys and the compound $CuCd_3$ forms a
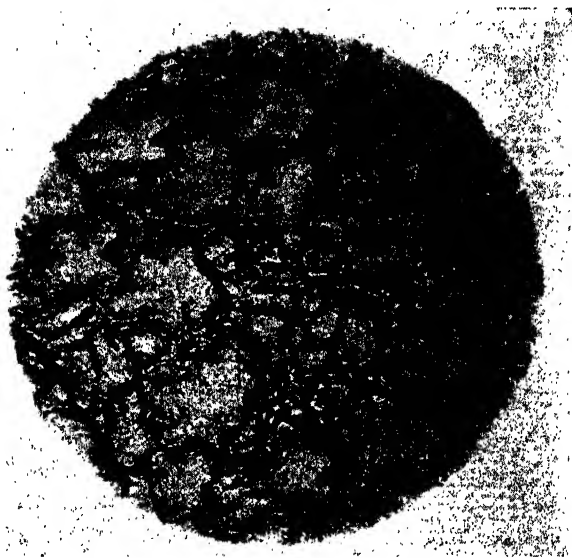
Fig. 34. -- Cadmium Base Bearing Alloy Containing 0.75% Ag and 0.5% Cu. (Etched with chromic acid.) Brinell 40. X200. (*Courtesy of Federal-Mogul Corporation.*)

eutectic mixture with the cadmium-rich solid solution. This relatively hard constituent freezes last, surrounding the crystals of the cadmium-rich solid solution as shown in Fig. 34, the microstructure of alloy CS50.

Nickel strengthens and hardens the cadmium-nickel bearing alloys. The 1.30% nickel type is the most widely used. Its microstructure, which resembles Fig. 33, contains cubical particles of the hard compound $NiCd_7$ in a matrix which consists of a plastic cadmium-rich eutectic ($Cd + NiCd_7$). The strength of the bond between this alloy and steel or bronze backing metals is known to be exceptionally high.

Zinc Base Bearing Alloys. — Zinc base bearing metals have been used for many years for special purposes, usually where surface speeds are high and the loads are steady and free from shock.   A typical composition: 85% Zn, 10% Cu, 5% Al, has a Brinell hardness of 130, which falls in the range of certain of the copper base bearing metals.   This composition has been used for the main spindle bearings of high speed lathes, for electric motor bearings, and other applications where rigidity and permanent alignment are important.   (The ternary equilibrium diagram for Zn-Cu-Al alloys has been determined by Hanson and Gaylor [16], thus the microstructure of this alloy may be interpreted from the diagram if desired.)

Zinc alloy die castings are used for some types of small electric motor frames which have no special bearing inserts.

Aluminum Alloys as Bearing Metals. — Aluminum base alloys have not been highly developed for specific application as bearing metals.   Nevertheless, certain aluminum alloy connecting rods have been successfully applied without the insertion of special bearing metals at the crankshaft end.   Bearing qualities are also an important consideration in aluminum piston alloys.   Many of these alloys are discussed in the following chapter.

Other Bearing Metals. — In Chapter VI the important copper base bearing alloys will be studied.   On page 131 the characteristics of the white metal bearings are compared with the copper base alloys and with pure silver for aircraft engine bearings.

Cast iron, an excellent bearing material for certain applications, is discussed in Chapter XIII, and the compositions of alloy steels used for ball and roller bearings are given in Chapter IX.

REFERENCES

1. " Modern Uses of Nonferrous Metals," Edited by C. H. Mathewson, Am. Inst. Mining and Met. Eng. (1935)
2. Bassett, H. N., " Bearing Metals and Alloys," Longmans, Green and Co., New York (1937)
3. Schumacher, E. E., and Hiers, G. O., " The Properties of Lead-Antimony Alloys," Metals Handbook, 1533, Am. Soc. for Metals, Cleveland (1939)
4. Fisher, V., " Babbitt Bearings," Metal Progress, **33,** 587 (1938)
5. Levy, A. H., " New Wire Permits Babbitt-Sprayed Bearings," Metallizer, **2,** 9 (Aug. 1934)

6. Ellis, O. W., "Physical Properties of Babbitt," Metal Progress, **29**, 56 (1936) — and — "Properties of Tin-Base Bearing Alloys," Metals Handbook, 1721, Am. Soc. for Metals, Cleveland (1939)

7. Ellis, O. W., "Structure of Tin-Antimony-Copper Alloys Used As Bearing Metals," Metals Handbook, 1709, Am. Soc. for Metals, Cleveland (1939)

8. Grant, L. E., "Bearing Metals of Lead Hardened by Alkali and Alkaline Earth Metals," Metals and Alloys, **5**, 161 and 191 (1934)

9. Hack, C. H., "A New Bearing Metal — A Lead Base Alloy," Metal Progress, **28**, 61 (Nov. 1935)

10. Briggs, J. Z., "Lead-Calcium Alloys for Storage Batteries," Metals and Alloys, **9**, 49 (1938)

11. Swift, L. L., "Microhardness of Bearing Alloys," Metals Technology, **5**, T. P. No. 966 (Sept. 1938)

12. Smart, C. F., "Cadmium-Silver-Copper Alloys for Engine Bearings," Trans. Am. Soc. for Metals, **25**, 571 (1937)

13. Smart, C. F., "Indium-treated Bearing Metals," Trans. Am. Inst. Mining and Met. Eng., **128**, 295 (1938)

14. Denham, A. F., "New Cadmium-silver-copper Bearing Alloys Developed by Federal-Mogul," Automotive Industries, **71**, 640 (Nov. 1934)

15. Swartz, C. E., and Phillips, A. J., "A Comparison of Certain White-Metal Bearing Alloys Particularly at Elevated Temperatures," Proc. Am. Soc. for Testing Mats., **33**, Part II, 416 (1933)

16. Hanson, D., and Gayler, M. L. V., "On the Constitution of Alloys of Aluminum, Copper, and Zinc," Jnl. Inst. of Metals, **34**, No. 2, 125 (1925)

CHAPTER IV

LIGHT ALLOYS

Properties of Aluminum and Magnesium. — Aluminum and magnesium are qualified by their low specific gravity, moderate cost, and good mechanical properties for consideration as basic materials in the light alloy field. Beryllium has some outstanding properties including its low density and high modulus of elasticity, but at the present stage of its metallurgical development it is much too expensive for large scale application, excepting as an alloying element in some of the heavy alloys.

The basic physical constants of these metals appear in Table I, page 4. The physical constants and mechanical properties of pure aluminum and magnesium are compared directly in Table III.

TABLE III.   COMPARISON OF PROPERTIES OF PURE ALUMINUM
AND MAGNESIUM

	Al	Mg	Mg/Al
1. Melting point, C°	657	651	—
2. Crystal structure	F.C.C.	Hex.C.P.	—
3. Specific gravity, g./cc.	2.70	1./4	0.64
4. Tensile strength, lb./sq. in.	8,500	28,000	3.3
5. Elongation, per cent in 2 in.	60	8	0.13
6. Modulus of elasticity, lb./sq. in.	10,000,000	6,250,000	0.62
7. Brinell hardness	16	35	—
8. Thermal conductivity (copper = 100)	54	40	0.74
9. Electrical conductivity (copper = 100)	63	37.7	0.60
10. Coef. of expansion (iron = 100)	202	216	1.07
11. Basic cost, ¢/lb. in Jan. 1939	20	31	—

(Items 4 to 6 are for annealed sheet metal of high purity.)

Many of these properties, especially items 4, 5, and 7, undergo important changes upon alloying and heat treatment. For example, the strength and hardness of aluminum may be increased to 65,000 lb. per sq. in. tensile strength and over 100 Brinell hardness.

Although the ductility will be reduced to around 20% elongation, the alloys are sufficiently ductile for most purposes; in fact, certain cold forming operations may be carried out on fully hardened material.   In like manner the magnesium alloys may be raised to 45,000 maximum tensile strength and 60 to 70 Brinell.   Only when relative densities are considered do the tensile properties of the magnesium alloys compare favorably with those of aluminum alloys. The hexagonal crystal structure of pure magnesium accounts for its comparative lack of plasticity, a characteristic which is somewhat improved in many of the commercial alloys.   Fortunately the magnesium alloys are readily hot worked at temperatures over 450° F. by rolling and extrusion methods, which makes possible the production of sheets, bars, and shapes.   Once fabricated the ductility is sufficient for ordinary applications.   The endurance characteristics of the magnesium alloys are equivalent to those of the aluminum alloys, even without taking into account their relative densities. This is an important consideration in the application of light alloys to machine parts in which rapidly alternating loads or vibrational stresses are encountered.

The thermal and electrical conductivities of magnesium alloys are somewhat less than those of aluminum alloys.   Both aluminum and magnesium have high coefficients of expansion (compared with steel); however, certain aluminum alloys have a considerably smaller coefficient than pure aluminum.

Comparison of Mechanical Properties of Light Alloys with Steels. — There are many methods of comparing the light alloys with steels for structural or machine parts.   The comparison may be on the basis of tensile, bending, or column members.   In any case the design of the actual sections will vary for the different materials used so as to utilize fully the properties available.   This partially invalidates a simple comparison such as that made in Table IV; however, these data give a rough approximation of relative efficiencies for structural work and other applications in which bending loads are important.

The values for yield strength divided by specific gravity, sometimes called strength-weight factors, show that high strength steel is somewhat " lighter " than aluminum alloys.   The exact values

TABLE IV.   COMPARISON OF ALLOYS FOR STRUCTURAL PURPOSES

ASSUMPTIONS

Material	Specific Gravity	Yield Strength lb./sq. in. (set = 0.2%)	Tensile Strength lb./sq. in.	Modulus of Elasticity lb./sq. in.	Yield Strength ÷ (S.G. × 44.3)
Mild steel	7.9	35 000	55 000	29 000 000	100
Low alloy steel	7.9	55 000	75 000	29 000 000	180
Alloy steel	7.9	140 000	180 000	29 000 000	400
Magnesium alloy	1.8	30 000	42 000	6 500 000	377
Aluminum alloy	2.8	40 000	60 000	10 300 000	323

COMPARISON OF BEAMS OF CONSTANT WIDTH AND VARIABLE DEPTH.

Load to develop yield strength stress varies as: Depth ^{2} × Yield Strength.
Elastic stiffness varies as: Depth ^{3} × Modulus of Elasticity.
Deflections vary inversely with elastic stiffness.

All values are relative, based on *mild steel = 100.*

Material	Depth of Beam	Weight of Beam	Load to Develop Yield Str.	Elastic Stiffness	Deflection
			Size of beam equal to that of mild steel		
Low alloy steel	100	100	157	100	100
Alloy steel	100	100	400	100	100
Magnesium alloy	100	23	86	22	447
Aluminum alloy	100	35	114	36	281
			Load to develop yield strength same as for mild steel		
Low alloy steel	80	80	100	51	196
Alloy steel	50	50	100	13	800
Magnesium alloy	109	25	100	29	350
Aluminum alloy	94	33	100	29	342
			Deflection same as for mild steel under a given safe load		
Low alloy steel	100	100	157	100	100
Alloy steel	100	100	400	100	100
Magnesium alloy	165	38	233	100	100
Aluminum alloy	141	50	227	100	100

for this relationship, and all others in the above table, depend, of course, on the original assumptions. Although yield strengths up to 55,000 lb. per sq. in. are available in certain aluminum alloys used for aircraft, guaranteed minimum specification values are much lower. Higher strength steels are also available, both in heat treated medium alloy types and in specially processed stainless grades; nevertheless, steels with lower yield strengths are very often used. Tuckerman [1] and Sutton [2] have made further comparisons of materials for aircraft construction.

For equivalent load carrying capacity as a beam, the weight of the alloy steel is one-half, the aluminum alloy one-third, and the magnesium alloy one-fourth the weight of the mild steel beam.

Importance of Modulus of Elasticity. — The alloy steel beam offers no weight advantage over mild steel as far as stiffness alone is concerned. The sections must be equal to those of ordinary mild steel if equal deflections are desired. Nevertheless, alloy steels and hardened carbon steels are widely used for leaf and coil springs because of their greater elastic strength. A high yield strength indicates ability to recover from higher working stresses, but not increased rigidity when stresses are within safe working limits.

The importance of modulus of elasticity in controlling the stiffness of beams and other structural members is obvious. The elastic moduli are practically unalterable by heat treatment or alloying, within the usual composition ranges, and high available strengths are sometimes not fully utilized because of deflection limitations. Aluminum alloy beams are designed with deeper sections than corresponding steel beams in order to offset this factor.

Low elastic moduli indicate a high capacity for absorbing energy imparted by shock or vibration in that greater elastic deformations result from a given stress on a given cross section.

Importance of Impact Toughness and Resistance to Repeated Stresses. — Failures by impact and by repeated stresses are further considerations which should not be neglected in practical applications. For example, if the beams are subjected to constantly varying stresses, such as would be produced by applying and removing a given load several times a second, the mild steel, magnesium alloy, and aluminum alloy beams of a given size might all fail by fatigue

after a large number of applications of the load.   Under the same conditions the alloy steel beam might last indefinitely because of its higher endurance limit.   Likewise the comparatively low notched-bar impact toughness of the light alloys should be considered in certain structural and machine designs.

There are many other considerations in the application of light alloys to machine and structural parts.   An excellent account may be found in the publication, " Service Characteristics of the Light Metals and Their Alloys." [3]   Detailed information on the entire field of aluminum alloys will be found in " The Aluminum Industry." [4]

Aluminum Alloys for Architectural Purposes. — The application of aluminum to window frames, doors, panels, and decorative trim in buildings depends upon its ability to retain attractive finishes in all types of atmosphere.   Light weight is of less importance than ease of fabrication in competition with stainless steels used for the same purposes.

Aluminum Heat Insulating Foil. — Another interesting and increasingly important application of aluminum depends upon its high reflectivity.   Aluminum foil reflects over 90% of the heat rays which strike it, even after it is well oxidized by atmospheric exposure.   The thin foil is usually corrugated and arranged in layers to provide air cells for further insulating value.   The advantages of aluminum foil include high insulating efficiency, light weight, and insensitivity to moisture and bacteria.   Successful applications range from refrigerators to steam turbines.

Requirements of Alloys for Pistons. — The use of light alloy castings and wrought shapes for aircraft, utensils, furniture, railway rolling stock, machine parts, as well as for ordinary structural work, has become too commonplace to require special consideration here. However, the requirements of alloys for internal combustion engine pistons are probably more exacting than those of any other light metal application and will be reviewed briefly in the following list:

(1) Light weight — to reduce bearing loads and engine vibration.   The inherent lightness of the aluminum alloy piston has been to a large extent equaled by the use of very thin

but strong sections by the makers of cast iron and steel pistons.

(2) High thermal conductivity — to facilitate dissipation of heat from the combustion chamber and prevent overheating of the head of the piston, thus avoiding preignition. This is a very definite advantage of aluminum pistons; one which becomes increasingly important as the size of the piston increases.

(3) Sufficient strength at operating temperatures.

(4) Wear resistance — to permit maintenance of the original fit for long periods of service and to reduce wear at the ring grooves.

(5) Low thermal expansion — to allow reasonably tight cold fits and prevent excessive wear at operating temperatures. The inherently high expansion of aluminum has been reduced in the modern piston alloys. Special design features are also employed to overcome high thermal expansion.

(6) Dimensional permanence — to avoid warping or growth at the usual operating temperatures. This is another feature which has been materially improved in aluminum piston alloys.

(7) Corrosion resistance — an important consideration when corrosive oils or gasolines are used.

Corrosion Resistance of Aluminum and Aluminum Alloys. — Aluminum surfaces are protected from atmospheric and certain other types of corrosion by the presence of an extremely thin coating of aluminum oxide which forms naturally in air. The alloys of aluminum are less effectively protected than the pure metal and are specially treated for many types of service. Aircraft sheets, for example, consist of one of the strongest alloys available coated on both sides with a thin layer of pure aluminum. This patented product, known as Alclad, is made by coating the alloy ingot with high purity aluminum, and reducing the duplex ingot by rolling. The bond has proved to be very efficient. The strong alloy is protected at exposed edges by galvanic action, the coating being electronega-

tive to the alloy. This is the same mechanism by which galvanized steel is protected by zinc. Zinc is also known to be electronegative to aluminum, which permits the use of galvanized steel wire for reinforcing strands of pure aluminum in electric conductor cables.

Certain of the strong aluminum alloys, especially those containing copper as the principal alloy addition, are subject to intergranular corrosion or preferential attack at the grain boundaries when put in service after improper heat treatment. In this case the best protection is proper heat treatment. (See page 85.) Intergranular attack is a particularly devastating type of corrosion, causing loss of strength and ductility and, in extreme cases, complete disintegration of the metal. This condition is not confined to any one class of alloys but has occurred in some one of its varied forms in widely different alloys including zinc die castings (page 96), pure copper (page 103), Monel metal, low carbon steel (page 186), and austenitic stainless steel (page 500).

Aluminum is resistant to attack by many chemicals, making it available for equipment and containers in the food and chemical industries for handling such materials as acetic acid, concentrated nitric acid, strong hydrogen peroxide, milk, and beer. Aluminum is not resistant to hydrochloric nor to many other non-oxidizing acids, nor to alkali.

Anodic Oxidation. — Other methods of protecting the aluminum alloys depend to a large extent upon building up the oxide coating artificially, usually followed by painting. The article to be treated is made the anode of a cell whose electrolyte may contain chromic acid, sulphuric acid, a nitrate, or some other oxidizing agent. There are several successful processes. Depending on the method used, the coatings will be more or less resistant to abrasion and will have good electrical resistance in addition to improved corrosion resistance. In most cases the anodic coating is somewhat porous, which makes it very suitable as a base for paints and dyes which materially increase the protection against corrosion.

Lanolin and other oils and greases are quite successful for coating the light alloys wherever the parts may be conveniently serviced.

Electroplating is a more costly means of protecting aluminum alloys. In many cases the principal purpose of the electrodeposited

coatings is to improve wear resistance. Zinc, chromium, and nickel are plated directly on aluminum under suitable conditions.

Certain alloys have been developed which are naturally more resistant to corrosion than others. These will be considered later.

Protection of Magnesium Alloys. — Magnesium alloys are less fortunate than aluminum in their natural resistance to corrosion. Surfaces tarnish to a gray color in ordinary atmosphere but become stable in this condition. The naturally formed surface film is less effective than that of aluminum. It is improved by the presence of a small amount of manganese which is added to all the commercial alloys. A sodium dichromate pickle produces a brassy finish well adapted to painting. Other treatments, including anodic oxidation, are being developed. As in the case of aluminum alloys, paints are applied over the treated surfaces to improve the protection.

The magnesium alloys are subject to rapid attack by alkaline salts and by most acids except concentrated hydrofluoric. Unlike the aluminum alloys, they are highly resistant to alkali. Crankcases and other engine castings for naval aircraft have been made of magnesium alloys, consequently it appears that the task of protecting such parts from brine corrosion is not hopeless.

Machinability of the Light Alloys. — While most fabricating operations are carried out more easily with aluminum than with magnesium alloys, due to the greater plasticity of the former, machining of aluminum is more difficult. The harder aluminum alloys and those in the fully heat treated state usually offer less difficulty than softer grades. Special cutting tools designed for cutting the light metals give best results. In general, such tools have greater rake and clearance angles than steel cutting tools and should be ground and polished to a smoother finish. Kerosene and mixtures of kerosene and lard oil are suitable cutting fluids. Accurate finishing cuts must be made without heating the work on account of the high thermal expansion. New free-machining aluminum alloys containing lead have been developed.[5]

Magnesium alloys are remarkably easy to machine. Using sharp tools without lubricants, high machining speeds may be attained. It has not been necessary to develop a special free-machining screw stock as in the case of many other types of alloys.

Welding of the Light Alloys. — Welding of the light alloys is commercially successful but has not yet reached the high state of efficiency now possible in steel welding practice. Some of the operating difficulties are caused by:

(1) Difficulty in removing the oxidized surface film and making an effective fusion.

(2) High thermal expansion.

(3) High thermal conductivity.

(4) Low melting temperatures.

(5) Weakening of cold rolled or heat treated alloys in the region of the weld.

(6) Difficulty in hardening the cast weld deposit by heat treatment, even though the composition is that of a potentially strong alloy.

(7) Hot-shortness of some compositions, or lack of sufficient strength in certain temperature ranges.

(8) Inferior corrosion resistance of the welded article.

Several of these conditions are, of course, common to the welding of many other alloys in addition to those of aluminum and magnesium.

The oxyhydrogen torch is recommended rather than the hotter oxyacetylene torch for gas welding aluminum sections under about ⅜ inch thick. A rod containing 5% silicon, balance mainly aluminum, is generally used for aluminum alloys because of certain special advantages including:

(1) Lower melting range than most of the alloys to be welded, and a wide enough solidification range to provide ample plasticity while the article is contracting during the cooling period.

(2) Low contraction on solidification.

(3) Freedom from hot-shortness.

(4) Better than the usual strength and resistance to corrosion in the welded state.

Aluminum and its alloys may be welded by the metallic and carbon arc processes. In the former process a heavily coated rod of the 5% silicon-aluminum type is used to advantage. The coating should be capable of forming a fusible slag for protection of the metal

from oxidation.   The high melting rate of the electrode and the high thermal conductivity of the stock necessitates rapid welding or preheating of the work in order to obtain proper fusion.

Electric resistance welds obviate many of the difficulties mentioned above and greatly extend the possibilities of fabricating both aluminum and magnesium alloys.   The high conductivity of aluminum alloys makes spot and seam welding by electrical resistance more difficult than for steels; however, equipment has been developed with which it is possible to make satisfactory welds.[6]

In general, magnesium alloys are more difficult to weld than aluminum; however, oxyacetylene and electric resistance methods are used successfully on certain grades.[7]

FIG. 35.— THE ALUMINUM-COPPER EQUILIBRIUM DIAGRAM.   THE SOLID SOLUTION θ IS PROBABLY THE COMPOUND $CuAl_2$ CONTAINING ABOUT 0.5% ALUMINUM IN SOLID SOLUTION.   (*Metals Handbook, 1939, p. 1222.*)

Soldering of the light alloys has not been highly successful because the joints tend to be low in strength and corrosion resistance.   The oxide coatings reduce adherence and their removal requires very powerful fluxes.   The aluminum solders are usually zinc-tin alloys.

The Aluminum-Copper System. — The aluminum-copper equilibrium diagram will be used to demonstrate the important principle of age hardening, which is the basis for the heat treatment of duralumin and many other alloys.

The important light alloys of aluminum and copper contain less than 15% copper.   According to Fig. 35 their microstructures consist of mixtures of α solid solution and eutectic, $(\alpha + \theta)$.   The eutectic, which, in this range, increases in amount with increase in copper content, is itself a fine mechanical mixture of α and the hard θ phase.

The solid solution theta (θ) was formerly believed to be the simple compound $CuAl_2$. Although the presence of the pure compound is now questioned, the term $CuAl_2$ is still widely used in referring to these alloys, and will be used in the following discussion. The alpha solid solution is of the simple substitutional type. The solubility increases from about 0.50% copper at room temperature to 5.65% copper at the eutectic temperature.

Left. Fig. 36. — High Purity Aluminum Alloy Containing 8% Copper, as Cast. (Etched with 0.5% HF in water.) X100. (*Courtesy of the Aluminum Company of America.*)

Right. Fig. 37. — High Purity Aluminum Alloy Containing 5% Copper. Cast Test Bar Heated 48 Hours at 544° C., Quenched in Water. (Etched with 0.5% HF in water.) X100. (*Courtesy of the Aluminum Company of America.*)

The microstructure of a cast, high purity aluminum alloy containing 8% copper is shown in Fig. 36. The eutectic constituent, being last to freeze, is found between the dendritic branches of the primary alpha solid solution. The $CuAl_2$ or θ particles of the eutectic increase the hardness of the alloy. Data are available for a similar high purity alloy containing 4.5% copper, showing a hardness as sand cast of 46 Brinell, compared with 16 Brinell for high purity wrought aluminum.[8]

The Solution Heat Treatment. — Alloys containing up to about 5.6% copper may be homogenized by heating to temperatures near the eutectic temperature of 548° C. for a sufficient time to dissolve the $CuAl_2$ (θ) phase and form a single alpha solid solution. Rapid cooling is necessary to preserve this condition at room temperature. The structure of a 5% copper cast alloy after the solution heat treatment is shown in Fig. 37. The $CuAl_2$ particles were entirely absent in other areas.[9] (Deeper etching would show alpha grain boundaries in this structure.)

The direct hardening effect of the $CuAl_2$ particles is apparently lost after they are dissolved in the alpha matrix. Nevertheless, the hardness of the alloy is increased by the presence of copper atoms in solid solution. Complete data for the 4.5% copper sand cast alloy, mentioned above, show that all the tensile properties are improved by the solution treatment.

TABLE V. EFFECT OF HEAT TREATMENT ON A HIGH PURITY ALUMINUM-COPPER ALLOY CONTAINING 4.5% COPPER

Treatment	Tensile Strength lb./sq. in.	Yield Point lb./sq. in.	Elongation in 2 in., %	Brinell Hardness
As Cast				
Aged 2 days at room temperature	20 140	8 820	7.5	46
Solution Heat Treatment				
Heated 40 hours at 540° C., quenched in water, tested immediately	35 820	17 440	20.7	62
Age Hardening Heat Treatment				
Heated 40 hours at 540° C., quenched in water, aged 2 days at room temperature ..	42 340	24 010	19.0	83

Courtesy of R. S. Archer and Z. Jeffries, Trans. Am. Inst. Mining and Met. Eng.[8]

Age Hardening. — The alpha solid solution obtained by quenching the 4.5% copper alloy from 540° C. to room temperature is definitely supersaturated with respect to copper, hence there is a strong tendency for the copper atoms to separate from the alpha crystals and reappear as $CuAl_2$ particles. This reaction is generally

believed to be responsible for the further change in properties in the direction of greater hardness and strength which is observed upon aging the quenched alloy at room temperature. (See Table V.) However, if $CuAl_2$ (θ) particles are precipitated at room temperature they are submicroscopic in size and only the planes of pre-
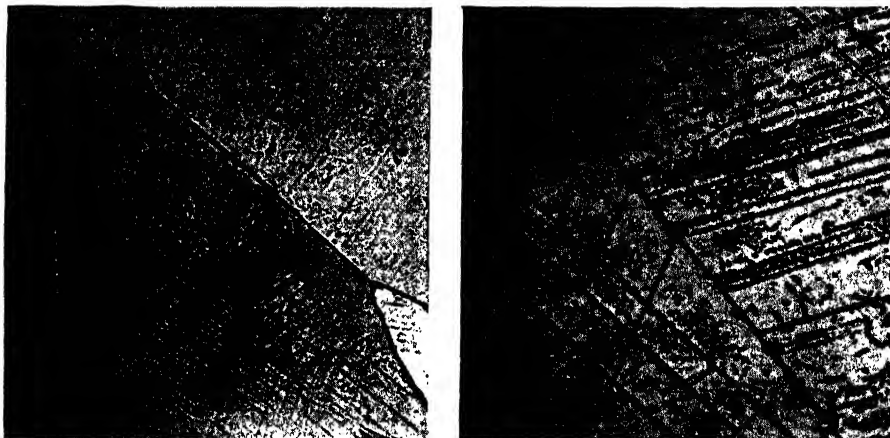
Left. FIG. 38. — HIGH PURITY ALUMINUM ALLOY SHEET CONTAINING 5.17% COPPER. HEATED 20 HOURS AT 540° C., QUENCHED IN WATER, AGED ½ HOUR AT 160° C. (Etched with 25% HNO_3 at 70° C., washed, and then etched in a solution containing: NaF, 0.5 g.; HNO_3, 1 cc.; HCl, 2 cc.; H_2O, 97 cc.) X100. (*Courtesy of Aluminum Company of America, Trans. Am. Inst. Mining and Met. Eng., Vol. 122, p. 284, 1936.*)

Right. FIG. 39. — SAME AS FIG. 38 EXCEPT AGED 2¼ HOURS AT 200°C. X500.

cipitation are evident even after applying the special etching procedure described in the caption for Fig. 38.[10]

Certain details of the precipitation mechanism have not as yet been conclusively established. It is quite likely, however, that considerable preliminary movement is necessary within the solid solution in order to form groups of atoms having a suitable arrangement for precipitation as the compound. The presence of these potential intermetallic compounds within the solid solution may, in itself, produce a considerable increase in resistance to deformation; in fact, the " knot formation " theory of the hardening of duralumin is based on this premise.[11]

Heating to elevated temperatures hastens precipitation of the

hardening particles. The results of treatments at two different temperatures are shown in Figs. 38 and 39. Although only the planes of precipitation (or "knot formation") are evident in Fig. 38, higher temperature and longer time of treatment has caused extensive precipitation of visible particles as shown in Fig. 39.

When $CuAl_2$ (θ) particles are precipitated from solid solution the copper atoms no longer exert their internal hardening and strengthening influence on the matrix; nevertheless, the presence of extremely large numbers of very small particles in certain crystallographic planes of the alpha crystals has an even greater hardening effect. This is the basis of the original theory of precipitation hardening as proposed twenty years ago.[12] Although several modifications, such as the "knot formation" theory, have been suggested from time to time, the most recent investigations on aluminum alloys have tended to substantiate the original conception. Nevertheless, authorities on age hardening are not agreed that the simple precipitation theory is universally applicable to age hardening alloys or capable of explaining certain complex reactions such as " double aging peaks " which may occur even in the aluminum alloys.[13]

Assuming that hardening is caused by actual precipitation, the mechanism by which the precipitated particles strengthen the matrix is still open to question. According to the slip interference theory proposed by Jeffries and Archer [14] the particles act as mechanical keys in potential slip planes. The most effective use of the available hardening material results when it is dispersed throughout the matrix in the form of particles of a critical size. According to Jeffries and Archer [14]: " The average size of particles which produces maximum hardness has been referred to as the *critical size,* and the hard substance is said to be in a condition of *critical dispersion.*" (To be considered further in the next section.) A second important hypothesis is based on distortion of the crystal lattice in the immediate vicinity of each precipitated particle. The resulting stiffening or " roughening " of slip planes is said to interfere with plastic deformation.

Aging at Elevated Temperature. — Certain commercial aluminum alloys age harden very slowly or not at all at room temperature

and must be reheated, usually to temperatures around 300° F., to develop high strength and hardness. Several hours may be required to produce maximum hardening at such temperatures, compared with several days at room temperature. At one time it was suggested that hardening by reheating to elevated temperatures be called artificial aging or precipitation hardening and that the term age hardening be reserved for the room temperature effect. Unfortunately this proposal has been quite generally disregarded and the terms age hardening and precipitation hardening are used almost interchangeably.

The effect of aging temperature is very well illustrated by Sykes' data on iron-tungsten alloys, reproduced in Fig. 40.[15] The lowest temperature at which the hardness curve reaches a maximum within the time available for treatment will generally give the highest hardness, cor-

FIG. 40. — TYPICAL AGE HARDENING BEHAVIOR ILLUSTRATED BY THE HARDNESS OF A QUENCHED 75% IRON-25% TUNGSTEN ALLOY AFTER REHEATING TO THE INDICATED TEMPERATURES. (*Courtesy of W. P. Sykes, E. S. Davenport, and E. C. Bain, Trans. Am. Soc. for Metals.*[15])

responding to the critical particle size. Higher temperatures are often used to obtain practically the same result in a much shorter time; however, the possibility of softening by overaging becomes greater at higher temperatures. Overaging is the result of coagulation of the hardening constituent into a relatively small number of large particles whose effect on the hardness and strength diminishes as the critical particle size is exceeded and the number of particles diminishes.

The effect of short periods of time at the higher aging temperatures is evident in curves by Davenport and Bain[15] on the aging of low carbon steel, shown in Fig. 41. These curves are qualitatively similar to those of Fig. 40 when the former are plotted to a uniform time scale. However, it is apparent from the examples

cited in this chapter that the effective temperature range for age hardening varies with the alloys treated.

Application of Age Hardening. — The theory of age hardening and its practical application in the production of strong aluminum alloys opened a period of development of other alloys having greater hardness and strength than previously thought possible. In a few years the nonferrous field was broadened so that it could
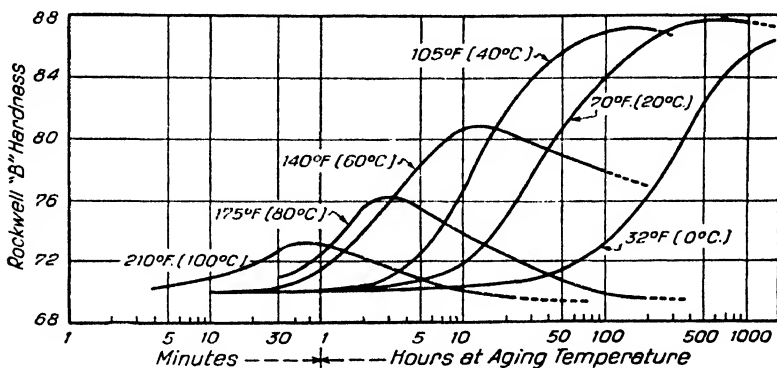
FIG. 41. — AGE HARDENING BEHAVIOR OF 0.06% CARBON STEEL QUENCHED FROM 1325% F. (*Courtesy of E. S. Davenport and E. C. Bain, Trans. Am. Soc. for Metals.*[15])

compete with irons and steels in many cases where competition was formerly impossible. Many new metals were developed which met requirements which no previous alloys, ferrous or nonferrous, could satisfy. Knowledge of the mechanism of age hardening has been useful in controlling certain undesirable aging effects encountered in low carbon steels. Precipitation plays a large part in the quench hardening of ordinary steels, according to one theory on this subject.

The future possibilities of precipitation hardening have been outlined by Merica [16], one of the early investigators in this field who was coauthor, in 1919, of the precipitation theory for the hardening of duralumin, and who proposed, in 1932, the conception of " knot formation " to explain reactions occurring in the early stages of the precipitation treatment.

The Aluminum-Mg_2Si System. — Another important hardening phase which is present in many of the strong aluminum alloys is the
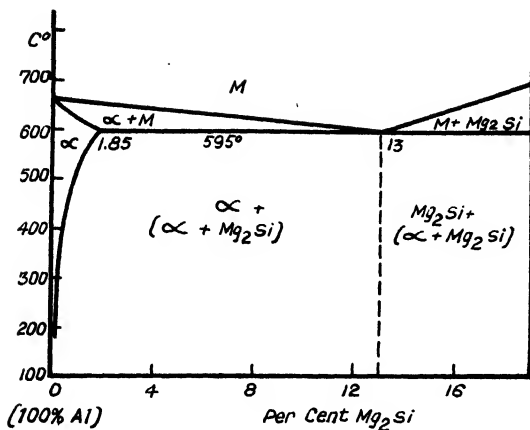
FIG. 42. — THE ALUMINUM-Mg₂Si EQUILIB-
RIUM DIAGRAM. (*Metals Handbook*, 1939, *p.*
1235.)

intermetallic compound Mg_2Si. Silicon and iron are present up to approximately 0.5% each in commercial aluminum. By adding proper amounts of magnesium, with or without further additions of silicon, this compound will form. Its equilibrium relations with aluminum are shown in Fig. 42. The similarity of this diagram and the aluminum-copper diagram, and the possibility of age hardening by means of Mg_2Si precipitation should be noted.

The Aluminum-Silicon System. — Silicon itself may be responsible for a small amount of age hardening as indicated by the solid solubility line in Fig. 43. However, the important aluminum-silicon casting alloys utilize the inherent hardness of the $(\alpha + Si)$ eutectic. These alloys may contain as high as 14% silicon, and according to the diagram their microstructures should consist of primary Si plus the $(\alpha + Si)$ eutectic constituent, as in Fig. 44.

It has been found that small additions of metallic sodium made to the melt just before casting prevent the formation of the massive primary silicon particles, lower the eutectic temperature, and refine the particle size of the eutectic as shown in Fig. 45. The mechanical
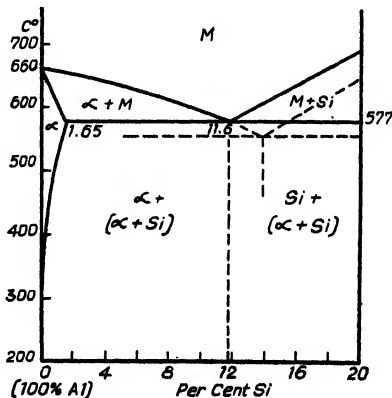
FIG. 43. — THE ALUMINUM-SILI-
CON DIAGRAM. (*Trans. Am. Inst.*
Mining and Met. Eng., Vol. 78, II,
1928, p. 169.)

properties are correspondingly improved. Rapid cooling, as in die casting, has a somewhat similar effect. The diagram indicated by the broken lines in Fig. 43 has been proposed to represent the conditions of solidification of the " modified " alloys. Approximately 15 or 16% silicon would be necessary to produce primary silicon particles in a fully modified alloy.

Iron, zinc, and several other elements are used in certain aluminum alloys. These elements often form special constituents in the alloys and participate in age hardening effects. Intensive studies have been made of many binary and ternary systems with high aluminum contents. This work has been of great value in the development of new alloys and in the treatment of those already existing. Summaries of the metallographic work on the aluminum alloys have been prepared by members of the Aluminum Company of America Research Laboratory.[17, 18]

Left.   FIG. 44. — CAST ALUMINUM ALLOY CONTAINING 14% SILICON, 0.03% IRON, 0.02% COPPER. (Unetched.) X100. (*Courtesy of Aluminum Company of America.*)

Right.   FIG. 45. — " MODIFIED " CAST ALUMINUM ALLOY CONTAINING 14% SILICON, 0.37% IRON, 0.02% COPPER. (Unetched.) X100. (*Courtesy of Aluminum Company of America.*)

TABLE VI.   APPROXIMATE COMPOSITIONS OF ALUMINUM SAND
CASTING ALLOYS

S.A.E. No.	Alcoa No.	PER CENT OF ALLOYING ELEMENTS — BALANCE ALUMINUM AND NORMAL IMPURITIES					
		Copper	Silicon	Magnesium	Iron	Nickel	Others
30	12	8.0	—	—	—	—	—
33 *	112	7.5	—	—	1.2	—	2.0 Zn
36	212	8.0	1.2	—	1.0	—	—
32	109	12.0	—	—	—	—	—
38	195	4.0	—	—	—	—	—
—	108	4.0	3.0	—	—	—	—
34 *	122	10.0	—	0.2	1.2	—	—
39 *	142	4.0	—	1.5	—	2.0	—
35 *	43	—	5.0	—	—	—	—
37	47	—	12.5	—	—	—	—
—	A334	3.0	4.0	0.3	—	—	—
322 *	355	1.3	5.0	0.5	—	—	—
—	A355	1.4	5.0	0.5	—	0.8	0.8 Mn
323	356	—	7.0	0.3	—	—	—
320	214	—	—	3.8	—	—	—
—	216	—	—	6.0	—	—	—
324	220	—	—	10.0	—	—	—
31A	645	2.5	—	—	1.5	—	11.0 Zn

* Also used for permanent mold castings.   (Additional data on compositions and properties of permanent mold alloys are given in " Alcoa Aluminum and Its Alloys.")

From " Alcoa Aluminum and Its Alloys," Aluminum Company of America, 1937.

Aluminum Casting Alloys. — The compositions and properties of important aluminum sand casting alloys are summarized in Tables VI and VII.   The outstanding characteristics and applications of the individual alloys will be mentioned briefly.   For further details refer to " Metals Handbook," " Society of Automotive Engineers Handbook," " Cast Metals Handbook," and technical data published by the Aluminum Company of America.

Heat Treatment of Aluminum Alloy Castings. — Many of the aluminum casting alloys are subject to improvement in hardness and strength by the application of solution and aging heat treatments.   In Table VII the symbol T4 refers to the solution heat treat-

TABLE VII. TYPICAL PROPERTIES OF SAND CAST ALUMINUM ALLOYS

S.A.E. No.	Alcoa No.	Yield Strength [a] lb./sq. in.	Tensile Strength [b] lb. sq./in. Typical	Tensile Strength [b] lb./sq. in. Minimum	Elongation in 2 in.,%	Brinell Hardness 500 kg. 10 mm.	Endurance Limit [c] lb./sq. in.
30 & 36	12 & 212	14 000	22 000	19 000	2.0	65	7 500
33	112	14 000	23 000	19 000	1.5	70	8 500
32	109	18 000	24 000	21 000	1.5	75	10 000
38	195–T4	16 000	31 000	29 000	8.5	65	6 000
38	195–T6	22 000	36 000	32 000	5.0	80	6 500
38	195–T62	31 000	40 000	36 000	2.0	95	7 000
—	108	14 000	21 000	19 000	2.0	55	8 500
34	122–T2	20 000	25 000	23 000	1.0	75	9 500
34	122–T61	30 000	36 000	30 000	1.0	100	—
39	142	24 000	28 000	23 000	1.0	80	8 000
39	142–T61	32 000	37 000	32 000	0.5	100	8 000
35	43	9 000	19 000	17 000	6.0	40	6 500
37	47 [d]	11 000	26 000	24 000	8.0	50	6 000
—	A334	16 000	25 000	22 000	2.0	65	—
322	355–T4	20 000	30 000	27 000	5.0	60	—
322	355–T6	25 000	35 000	32 000	3.5	80	—
323	356–T4	16 000	28 000	26 000	6.0	55	—
323	356–T6	22 000	32 000	30 000	4.0	70	8 000
320	214	12 000	25 000	22 000	9.0	50	5 500
—	216	15 000	27 000	25 000	6.0	60	—
324	220–T4	25 000	45 000	42 000	14.0	75	7 000
31A	645	20 000	29 000	25 000	4.0	70	7 500

[a] Set = 0.2%. Modulus of elasticity is approximately 10 300 000 lb./sq. in.

[b] Tension values determined from standard half inch diameter tensile test specimens individually cast in green sand molds and tested without machining off the surface.

[c] Based on withstanding 500 000 000 cycles of completely reversed stress, using the R. R. Moore type of machine and specimen.

[d] Properties of this alloy obtained by special foundry practice, called " modification." (Data for additional heat treatments and properties are given in the original table.)

From " Alcoa Aluminum and Its Alloys," Aluminum Company of America, 1937.

ment. Dendritic segregation in the castings makes it necessary to heat for several hours to approach a homogeneous structure in the solution heat treatment. Aging treatments are indicated by T6 and other similar symbols in Table VII. Increase in hardness and strength by aging is generally accompanied by considerable reduction in ductility in the cast alloys, whereas comparable wrought alloys not only develop higher strength, but also much higher ductility in the fully aged condition.

Permanent mold and die castings, and to a lesser extent sand cast-

ings, often retain sufficient hardening agents in solid solution to re-
spond to aging treatments applied directly after casting. In this way
intricate castings can be hardened and strengthened without intro-
ducing quenching stresses in a solution heat treatment. The tend-
ency to " grow " or increase in dimensions at elevated operating
temperatures is reduced by this treatment. The solution heat treat-
ment is necessary, however, to improve ductility and shock resist-
ance.

High Copper Casting Alloys. — The aluminum-copper alloys
were used almost exclusively in the early days of the aluminum cast-
ing industry. The alloys containing about 8% copper are still widely
used for general purposes where the requirements for corrosion re-
sistance, strength, and castability are not great. The 12% copper
alloy has less shock resistance but is suitable for manifolds, pump
housings, and other castings requiring pressure-tightness.

The 4% copper alloy, No. 195, develops good physical properties
upon heat treatment. It is apparent, however, from the data in
Table VII that the increased hardness and strength produced by
heat treatment are accompanied by a considerable reduction in duc-
tility. A disadvantage of this alloy is its hot-shortness, which makes
it difficult to obtain pressure-tight castings of intricate design. It
is used for general engineering purposes including automotive, air-
craft, and electrical fittings, machine tool parts, crankcases and
other engine castings.

Alcoa Nos. 122 and 142 are essentially permanent mold casting
alloys used mainly for internal combustion engine pistons. Alloy
No. 142, otherwise known as Y alloy, contains nickel, which im-
proves its strength and resistance to growth at elevated tempera-
tures.

The copper type alloys are, in general, comparatively easy to
machine.

High Silicon Casting Alloys. — The 5% silicon alloy is one of a
series of high silicon alloys which have become increasingly impor-
tant because of their good corrosion resistance, good weldability,
low specific gravity, comparatively low thermal expansion, and high
thermal conductivity. They are relatively easy to cast because of
high fluidity and freedom from hot-shortness. The strength of this
silicon alloy is not exceptional, but it has better ductility and impact

toughness than the high copper types. It is used for architectural castings, outboard motors and marine fittings, manifolds, water jackets, and other castings requiring pressure-tightness and corrosion resistance.

The 12% silicon " modified " alloy has higher strength and very good ductility. It has been widely used for intricate pressure-tight castings requiring good mechanical properties in the cast condition without further treatment.

Alcoa Nos. 355 and 356 contain small amounts of magnesium which make it possible to heat treat for higher strength. No. 355 was developed especially for heat treated cylinder heads for water cooled engines. Additions of nickel and manganese improve the elevated temperature properties of A355, making it suitable for air cooled cylinder heads of aircraft engines. Alloy No. 356 has been used for crankcases and other engine, pump, and compressor castings. These alloys are replacing the copper type aluminum alloys in many applications of the type mentioned.

High Magnesium Casting Alloys. — Aluminum alloys containing high magnesium contents are recognized as being the most resistant to corrosion, especially in salt laden atmospheres and brine solutions, of any of the available light alloys. Although their casting characteristics are unfavorable, requiring special foundry practice to prevent oxidation or drossing, the mechanical properties of sound castings are highly satisfactory, especially in the case of the 10% magnesium alloy after the solution heat treatment. The densities of the magnesium type aluminum castings are the lowest obtainable in aluminum base alloys.

The 3.75% magnesium alloy is used for cast cooking utensils, dairy and plumbing equipment, pipe fittings and valves, ornamental castings, and other applications where corrosion resistance is one of the principal requirements. The 10% magnesium alloy, whose specific gravity is about 10% less than that of an 8% copper alloy, is used for aircraft fittings and other castings requiring maximum strength, ductility, and corrosion resistance.

High Zinc Casting Alloys. — The high zinc alloys are not used extensively in this country. The mechanical properties are satisfactory at room temperature but are said to be poor at elevated temperatures. Corrosion resistance is relatively poor.

TABLE VIII. NOMINAL COMPOSITIONS OF WROUGHT ALUMINUM ALLOYS

Per Cent of Alloying Elements — Balance Aluminum and Normal Impurities

Alcoa No.	Copper	Silicon	Man-ganese	Mag-nesium	Zinc	Nickel	Chro-mium	Lead	Bis-muth
2S	—	—	—	—	—	—	– –	—	—
3S	—	—	1.2	—	—	—	—	—	—
4S	—	—	1.2	1.0	—	—	—	—	—
11S	5.5	—	—	—	—	—	—	0.5	0.5
14S	4.4	0.8	0.8	0.4	—	—	—	—	—
17S	4.0	—	0.5	0.5	—	—	—	—	—
A17S	2.5	—	—	0.3	—	—	—	—	—
18S	4.0	—	—	0.5	—	2.0	—	—	—
24S	4.4	—	0.5	1.5	—	—	—	—	—
25S	4.5	0.8	0.8	—	—	—	—	—	—
32S	0.8	12.0	—	1.0	—	0.8	—	—	—
51S	—	1.0	—	0.6	—	—	—	—	—
A51S	—	1.0	—	0.6	—	—	0.25	—	—
52S	—	—	—	2.5	—	—	0.25	—	—
53S	—	0.7	—	1.3	—	—	0.25	—	—
56S	—	—	0.1	5.2	—	—	0.1	—	—
70S	1.0	—	0.7	0.4	10.0	—	—	—	—

From " Alcoa Aluminum and Its Alloys," Aluminum Company of America, 1937.

TABLE IX. TYPICAL PROPERTIES OF WROUGHT ALUMINUM ALLOYS

Alcoa No.	Yield Strength a lb./sq. in.	Tensile Strength lb./sq. in.	Elongation in 2 in., % b		Brinell Hardness 500 kg. load 10 mm. ball	Shearing Strength c lb./sq. in.	Endurance Limit d lb./sq. in.
			Sheet	Bar			
2S–O	5 000	13 000	35	45	23	9 500	5 000
2S–¼H	13 000	15 000	12	25	28	10 000	6 000
2S–½H	14 000	17 000	9	20	32	11 000	7 000
2S–¾H	17 000	20 000	6	17	38	12 000	8 000
2S–H	21 000	24 000	5	15	44	13 000	8 500
3S–O	6 000	16 000	30	40	28	11 000	7 000
3S–¼H	15 000	18 000	10	20	35	12 000	8 000
3S–½H	18 000	21 000	8	16	40	14 000	9 000
3S–¾H	21 000	25 000	5	14	47	15 000	9 500
3S–H	25 000	29 000	4	10	55	16 000	10 000

TABLE IX. (*Continued*)

Alcoa No.	Yield Strength[a] lb./sq. in.	Tensile Strength lb./sq. in.	Elongation in 2 in., %[b] Sheet	Bar	Brinell Hardness 500 kg. load 10 mm. ball	Shearing Strength[c] lb./sq. in.	Endurance Limit[d] lb./sq. in.
4S–O	10 000	26 000	20	25	45	16 000	14 000
4S–¼H	22 000	31 000	10	17	52	17 000	14 500
4S–½H	27 000	34 000	9	12	63	18 000	15 000
4S–¾H	31 000	37 000	5	9	70	20 000	15 500
4S–H	34 000	40 000	5	6	77	21 000	16 000
11S–T3	42 000	49 000	—	14	95	30 000	12 500
17S–O	10 000	26 000	20	22	45	18 000	11 000
17S–T	37 000	60 000	20	22	100	36 000	15 000
17S–RT	47 000	65 000	13	—	110	38 000	—
Alclad 17S–T	33 000	56 000	18	—	—	32 000	—
Alclad 17S–RT	40 000	57 000	11	—	—	32 000	—
A17S–O	8 000	22 000	24	27	38	15 000	—
A17S–T	24 000	43 000	24	27	70	26 000	13 500
24S–O	10 000	26 000	20	22	42	18 000	12 000
24S–T	44 000	68 000	19	22	105	41 000	16 000
24S–RT	55 000	70 000	13	—	116	42 000	—
Alclad 24S–T	41 000	62 000	18	—	—	40 000	—
Alclad 24S–RT	50 000	66 000	11	—	—	41 000	—
51S–O	6 000	16 000	30	35	28	11 000	6 500
51S–W	20 000	35 000	24	30	64	24 000	10 500
51S–T	40 000	48 000	14	16	95	30 000	10 500
52S–O	14 000	29 000	25	30	45	18 000	17 000
52S–¼H	26 000	34 000	12	18	62	20 000	18 000
52S–½H	29 000	37 000	10	14	67	21 000	19 000
52S–¾H	34 000	39 000	8	10	74	23 000	20 000
52S–H	36 000	41 000	7	8	85	24 000	20 500
53S–O	7 000	16 000	25	35	26	11 000	7 500
53S–W	20 000	33 000	22	30	65	20 000	10 000
53S–T	33 000	39 000	14	20	80	24 000	11 000

[a] Set = 0.2%. Modulus of elasticity is approximately 10 300 000 lb. per sq. in.

[b] Values are for 1/16 inch thick sheets and standard 0.505 inch diameter specimens.

[c] Single-shear strength values obtained from double-shear tests.

[d] Based on withstanding 500 000 000 cycles of completely reversed stress, using the R. R. Moore type of machine and specimen.

From " Alcoa Aluminum and Its Alloys," Aluminum Company of America, 1937.

Wrought Aluminum Alloys. — The wrought aluminum alloys of Tables VIII to XI constitute the most important group from the standpoint of actual tonnage of products. The requirements and uses of the rolled, drawn, extruded, and forged alloys are naturally quite different from those of castings. Because of the beneficial effect of mechanical working in homogenizing the structure, the heat treatments are more effective and the mechanical properties, especially ductility and toughness, are superior to those of castings. Furthermore, the total percentage of alloying agents needed to provide the necessary strength is reduced.

Strain-Hardened Alloys Used Without Heat Treatment. — Commercially pure wrought aluminum, Alcoa 2S, contains approximately 99.0% aluminum, the balance being mainly silicon and iron, elements which are present in all of the alloys even though not specifically indicated in the recorded compositions. This composition satisfies the requirements of lightness, corrosion resistance, ease of fabrication, and low cost; consequently it is used, usually as sheet, plate, tubing, bars, extruded sections, or wire where high strength is not needed.

In making these products the cast ingots are reduced by hot working, often followed by cold working to final dimensions. Finished sheet, tubing, and wire may be purchased in the strain-hardened condition as quarter hard, one-half hard, etc.: designated ¼ H, ½ H, ¾ H, H. When cold forming operations are to be carried out, the stock must not be too hard or annealing may be required. However, the hardest grade which will withstand the drawing, cupping, spinning, or riveting operations will give the strongest product. If annealing is required, air cooling from 650° F. will be adequate to remove all work hardness. There is no necessity for slow cooling since there are no significant phase changes in this commercially pure metal. Annealed aluminum alloys are represented by the letter " O " as in 2S–O.

Rather than develop strength entirely by cold working, the alloys 3S and 4S may be used. These compositions have greater inherent hardness, available without heat or mechanical treatment. Alloy 3S is only slightly less workable and slightly less resistant to corrosion than 2S. It retains its strength at elevated temperatures to

a greater extent than 2S because manganese raises the recrystalliza-
tion temperature. It is extensively used for cold formed cooking
utensils, for roofing, doors, panels, and other building purposes,
and for furniture. Commercial aluminum and the low alloy types
are more readily welded than the stronger alloys.

Properties intermediate between those of 3S and those of the
heat treated alloys are available in the strain-hardening alloys 4S
and 52S. Comparing the two alloys, 52S has better forming prop-

TABLE X. CONDITIONS FOR HEAT TREATMENT OF WROUGHT
ALUMINUM ALLOYS

Alcoa No.	Solution Heat Treatment			Precipitation Heat Treatment		
	Temperature, ° F.	Quenching Medium	Temper Designation	Temperature, ° F.	Time of Aging	Temper Designation
17S	930–950	cold water	—	room	4 days	17S–T
A17S	930–950	cold water	—	room	4 days	A17S–T
24S	910–930	cold water	—	room	4 days	24S–T
51S	960–980	cold water	51S–W	315–325	18 hours	51S–T
53S	960–980	cold water	53S–W	315–325	18 hours	53S–T

From " Alcoa Aluminum and Its Alloys," Aluminum Company of America, 1937.

erties, higher strength and endurance limit, and better corrosion
resistance. It is used in the form of sheet and plate in the construc-
tion of equipment for marine, railway, and motorbus service, and
for architectural purposes.

Cold drawn wire of alloy 56S develops the highest strength ob-
tainable in the strain-hardening alloys. However, this composition
is not used for sheets or other wrought forms.

Heat Treatment of Wrought Alloys. —The conditions for heat
treatment of certain wrought alloys are given in Table X.

Alloys 17S, A17S, and 24S age harden at room temperature after
the solution heat treatment. The letter " T " is used to indicate the
hardened condition.

Alloys 51S, 53S, and some of the forging alloys of Table XI, do
not harden fully at room temperature after the solution heat treat-
ment (the letter " W " is used to designate this condition), but
harden by precipitation at elevated temperatures.

The strength of the heat treated alloys may be further increased by small amounts of cold work, indicated in the alloy symbols by the letters " RT." (See Table IX.) The improvement in strength by strain-hardening is limited by the reduction in ductility which takes place upon cold working.

Heating for the solution treatment is best accomplished in a fused nitrate salt bath, in which the heating time will vary from 10 to 60 minutes depending on the charge. Heating in air requires a longer time. The temperatures must be closely controlled and uniform throughout the charge in all heat treating operations.

Precipitation heat treatment at elevated temperatures, as well as many of the compositions of aluminum alloys, is patented.

Aging at room temperature progresses rapidly after quenching, more than 90% of the maximum properties being obtained in one day. Steam heated ovens and electric furnaces are used for aging at elevated temperatures. Aging may be delayed by storing at a low temperature. The superior forming properties of newly quenched 17S rivets may be retained until ready for use by storing in a dry-ice container.

Annealing of hardened alloys may be accomplished by heating to 600° to 800° F. and cooling slowly to coagulate the hardening constituents into large particles.

Heat Treated Wrought Aluminum Alloys. — The strong, heat treated aluminum alloys used for structural purposes and forgings originated in Germany, where they were used as early as 1905. (The history of aluminum alloy development will be found in " The Aluminum Industry." [4]) The alloy 17S, which ages at room temperature, is the type which was used in the early days of the industry and is still the most important composition for general structural purposes. The term " duralumin " is often used in referring to this alloy and to others of a similar nature. The modification A17S was developed to withstand more severe forming operations.

The alloy 24S is a recent development for applications requiring higher strength. It has largely replaced 17S for aircraft construction, particularly for highly stressed sheets. Alclad 24S–RT is used in very thin gages because of its high strength and corrosion resistance.

Although it is possible to harden 17S, 24S, and similar alloys by precipitation at elevated temperatures for relatively short periods of time, compared with room temperature aging, this treatment is to be avoided because it greatly increases the susceptibility of the aluminum-copper type alloys to intergranular corrosion.

TABLE XI. MECHANICAL PROPERTIES SPECIFICATIONS FOR ALU-MINUM ALLOY FORGINGS

Alcoa No.	Tensile Strength lb./sq. in. Minimum	Yield Strength lb./sq. in. (Set = 0.2%) Minimum	Elongation in 2 in., % Minimum	Brinell Hardness 500 kg. load 10 mm. ball Minimum
11S–T	55 000	34 000	12	90
14S–T	65 000	50 000	10	130
17S–T	55 000	30 000	16	100
18S–T	55 000	35 000	10	100
25S–T	55 000	30 000	16	100
32S–T	52 000	40 000	5	115
A51S–T	44 000	34 000	14	90
53S–T	36 000	30 000	16	75
70S–T	50 000	40 000	16	85

Applies to forgings up to 4 inches in diameter or thickness. Long axis of test specimen taken parallel to direction of grain flow.

From " Alcoa Aluminum and Its Alloys," Aluminum Company of America, 1937. (Additional data are given in the original table.)

The copper-free alloy 51S is hardened by precipitation of Mg_2Si particles. It is readily formed in the quenched condition, designated 51S–W, and may then be heated to the aging temperature for hardening.

Alloy 53S was developed for architectural and structural applications requiring moderately high strength with atmospheric corrosion resistance approaching that of 2S. It is similar to the strain-hardening alloy 52S and has likewise found wide application.

The free-machining alloy, 11S, is generally used in bar form as screw machine stock. The products may be heat treated for high strength.

Aluminum Forging Alloys. — The properties of heat treated forgings are given in Table XI. (Note that in this case minimum

specifications are given rather than typical values.) The most widely used forging alloy is 25S which has mechanical properties similar to 17S and better hot working characteristics. Among its important applications are propellers and connecting rods for aircraft engines. Alloy 14S has forging characteristics similar to 17S but develops higher strength and hardness upon heat treatment. Intricate forgings which are difficult to make with the strongest alloys can be most readily forged using alloy A51S.

Parts requiring heat resistance, such as aircraft pistons, are forged from 18S and 32S. (Howell and Paul [19] have reported the elevated temperature properties of several aluminum alloys.) The high silicon type, 32S, has the advantage of a coefficient of expansion about 13% lower than that of the high copper type of permanent mold casting alloys used for pistons.

Forgings which must meet severe corrosion requirements may be made of 53S.

The high zinc alloy, 70S, is a lower cost composition of the type used mainly in Europe.

Aluminum Electric Conductor Cables. — About 15% of the aluminum used in this country is made into electric conductor cables which compete with copper for long span, high voltage construction. The electrical conductivity of hard drawn aluminum is about 61% that of copper on a volume basis and over twice as great on the basis of equal weights. Both aluminum and copper wires are used in a cold drawn condition in order to increase the strength with only a small sacrifice of conductivity. Hard drawn copper, for example, has a conductivity of about 97.5% of the copper standard, compared with about 101% for annealed electrolytic copper. The decrease of conductivity of cold drawn compared with annealed aluminum is comparable. The strength of the drawn wire increases as the diameter decreases, ranging from about 23–30,000 lb. per sq. in. for aluminum and 49–67,000 lb. per sq. in. for copper. In order to compensate for their comparatively low strength, the aluminum wires are fabricated into a composite cable having a high strength steel wire or cable as a core. The steel core takes more than one-half of the load. It is zinc coated to increase its resistance to corrosion.

Aluminum cables are lighter and stronger than electrically equivalent copper cables of conventional design, and have greater outside diameters. The lower corona loss experienced with aluminum conductors is the result of the large diameters of the cables. Hollow copper conductor cables also take advantage of this effect.

The higher cost of aluminum cables is only one of the economic factors involved; for example, longer spans are possible which reduces the investment in steel towers.

TABLE XII.   TYPICAL COMPOSITIONS OF MAGNESIUM ALLOYS

	Designation		Per Cent Composition (Balance Magnesium)			Charac-teristic Use
A.S.T.M. (B80–38T)	Dow Chemical Co.	American Magnesium Corp.	Al	Mn	Zn	
2	G	AM240	10.0	0.1	—	Castings
3	B	AM246	12.0	0.1	—	"
4	H	AM265	6.0	0.2	3.0	"
14	P	—	10.0	0.1	1.0	"
11	M	AM3S	—	1.5	—	Rolled sheet
6	F	AM53S	4.0	0.3	—	and plate
8	J	AM57S	6.5	0.3	0.75	Press forgings,
9	O	AM58S	8.5	0.2	0.5	extruded rod and shapes

From Metals Handbook, 1939, pages 1566 and 1587.

In European countries considerable progress has been made in applying a high strength aluminum alloy similar to 51S without using a steel core.

Magnesium Base Alloys. — The compositions and properties of typical magnesium base alloys are given in Tables XII and XIII. Two outstanding properties not evident from these data are extremely light weight, about two-thirds that of aluminum alloys, and excellent machinability.

Aluminum is the principal alloying element used to improve the mechanical properties; however, zinc is also added to most of the newer alloys because it improves both the mechanical properties and resistance to corrosion. A small amount of manganese also improves corrosion resistance.

TABLE XIII.   TYPICAL PROPERTIES OF MAGNESIUM ALLOYS

Dow Chemical Co.	American Magnesium Corp.	Condition	Yield Strength lb./sq. in.	Tensile Strength lb./sq. in.	Elongation in 2 in. %	Brinell Hardness	Shear Strength lb./sq. in.	Endurance Limit lb./sq. in.
A	AM241–T4	Solution heat treatment	11 000	33 000	10.0	48	—	7 500
G	AM240C	Sand cast	11 000	20 000	2.0	52	—	8 000
G	AM240–T4	Solution heat treatment	12 000	35 000	9.0	52	—	11 000
G	AM240–T6	Heat treated and aged	16 000	35 000	4.0	60	—	8 000
G	AM240–T61	Heat treated and aged	19 000	36 000	1.0	69	—	8 000
B	AM246C	Sand cast	14 000	19 000	0.5	65	—	6 000
B	AM246–T6	Heat treated and aged	20 000	32 000	0.5	85	—	7 000
H	AM265C	Sand cast	11 000	27 000	6.0	48	—	11 000
H	AM265–T4	Solution heat treatment	12 000	35 000	9.0	51	—	11 000
H	AM265–T6	Heat treated and aged	18 000	37 000	4.0	69	—	9 000
M	AM3S	Extruded and stretched	26 000	35 000	7	40	16 000	—
F	AM53S	Extruded and stretched	30 000	40 000	17	50	19 000	12 000
J	AM57S	Extruded and stretched	33 000	44 000	15	55	20 500	15 000
O	AM58S	Extruded and stretched	36 000	47 000	13	60	21 500	16 000
M	AM3S–T	Soft (heat treated)	15 000	32 000	16	40	17 500	8 000
M	AM3S–RT	Hard (rolled after heat treated)	29 000	36 000	10	52	—	—
F	AM53S–O	Annealed sheet	22 000	36 000	18	50	—	—
F	AM53S–H	Hard rolled sheet	35 000	44 000	10	60	—	—
J	AM57S	Press forging	26 000	42 000	8	55	21 000	14 000
O	AM58S	Press forging	30 000	45 000	6	60	22 000	15 000

Data taken mainly from " Magnesium Alloys," Aluminum Company of America, Nov. 16, 1936.

Aluminum and zinc are both soluble in solid magnesium to a limited extent, making it possible to harden certain of the alloys by solution and aging heat treatments comparable to those applied to the aluminum base alloys. Equilibrium diagrams and other detailed information on these alloys will be found in " Metals Handbook."

Magnesium Alloy Castings. — Magnesium alloys require special foundry practice, including a protective flux during melting and pouring, a specially treated molding sand to prevent reaction with the metal, and large gates and risers to compensate for the low pressure head.[20] Under proper conditions sound castings are produced by sand, permanent mold, and die casting methods.

The typical compositions listed in Table XII contain 9 to 12% of alloying elements which is the optimum range for good casting characteristics. The high mechanical properties of the 6% aluminum-3% zinc type, combined with its superior corrosion resistance have made it the leading casting alloy. The 12% aluminum alloy develops the highest hardness upon heat treatment but is low in ductility and impact toughness. It

FIG. 46. — TORQUE-ANGULAR DEFORMATION CURVES FOR COLD ROLLED ALCOA 51S AND EXTRUDED DOWMETAL A.

is being replaced by A.S.T.M. No. 14 for castings requiring highest yield strength and hardness and not subject to shock.

Magnesium alloy castings are becoming increasingly popular for household appliances, foundry equipment, portable tools, aircraft
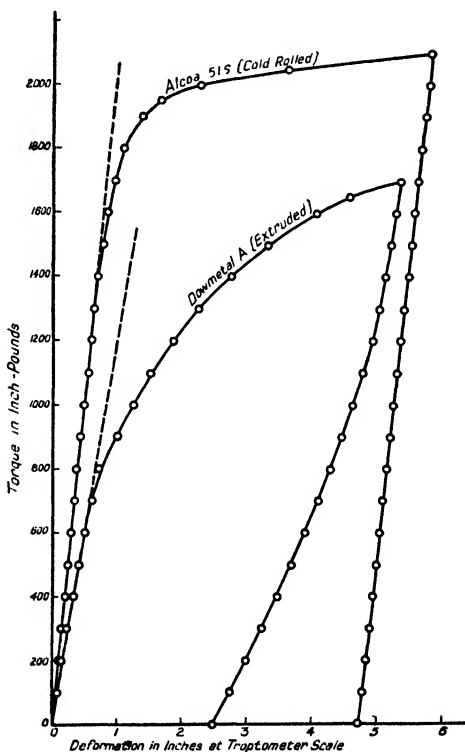
gear and engine castings, high speed machinery, and other applications where extreme lightness is desired.

Wrought Magnesium Alloys. — The alloy containing 1.5% manganese as the only hardening element is used for sheets and extruded bar where lightness, formability, and corrosion resistance are the essential factors as in aircraft gasoline tanks, engine cowlings, and fuel line fittings. The 4% aluminum alloy is specified for sheets, bars, plates and extruded shapes where higher strength and ductility is required. The 6.5% and 8.5% aluminum alloys are used for extrusion and press forging. Most extruded sections are made of the 6.5% aluminum alloy. Alloys in which tin is the principal addition are available for hammer forgings. Aircraft propellers, crankcases, nose pieces, wheels, and other parts have been forged using magnesium alloys. The possibilities of magnesium alloy forgings for highly stressed parts subject to fatigue failure are exemplified by their successful application, in Europe, as aircraft propellers.

An interesting characteristic of the magnesium alloys was observed while making tensile and torsional tests of bar stock. Typical torsion curves for magnesium and aluminum alloys are plotted in Fig. 46. Troptometers having a 13 inch radius were used on 8½ inch gage lengths to measure the angular displacements upon loading and unloading ⅞ inch diameter bars. The " release " curve for the aluminum alloy (as well as for several other alloys tested) is practically straight and parallel to the original modulus line, while that of the magnesium alloy curves markedly towards the origin, indicating a high " recovery " after plastic deformation and a high capacity for absorbing energy upon loading and unloading.

REFERENCES

1. Tuckerman, L. B., " Aircraft Materials and Testing," Proc. Am. Soc. for Testing Mats., **35,** Part II, 3 (1935)
2. Sutton, W. L., " Aluminum Alloys vs. Stainless Steels for Aircraft," Metal Progress, **31,** 46 (1937)
3. " Service Characteristics of the Light Metals and Their Alloys," Proc. Am. Soc. for Testing Mats., **34,** Part I, 277 (1934)
4. Edwards, J. D., Frary, F. C., and Jeffries, Z., " The Aluminum Industry — Aluminum Products and Their Fabrication," McGraw-Hill Book Company, New York (1930)
5. Kempf, L. W., and Hartwell, A., " Screw Machine Performance on New Aluminum Alloy," Metal Progress, **30,** 34 (Dec. 1936)

6. Bohn, D. I., " Spot and Seam Welding of Aluminum Alloy Sheet," Metal Progress, **32**, 53 (1937)

7. "Welding of Metals," Metals Handbook, 220, Am. Soc. for Metals, Cleveland (1939)

8. Archer, R. S., and Jeffries, Z., " Aluminum Castings of High Strength," Trans. Am. Inst. Mining and Met. Eng., Inst. of Metals Div., 35 (1927)

9. " Structure of Aluminum-Copper Alloys," Metal Progress, **18**, 81 (Oct. 1930)

10. Fink, W. L., and Smith, D. W., " Age-hardening of Aluminum Alloys, I-Aluminum-copper Alloy," Trans. Am. Inst. Mining and Met. Eng., **122**, 284 (1936)

11. Merica, P. D., Waltenberg, R. G., and Scott, H., " Heat Treatment of Duralumin," U. S. Bur. of Stds., Scientific Papers, **15**, 271 (No. 347) (1919)

12. Sykes, W. P., " New Light on Old Metallurgical Problems," Mining and Metallurgy, **20**, 54 (1939)

13. Jeffries, Z., and Archer, R. S., " The Science of Metals," McGraw-Hill Book Company, New York (1924)

14. Davenport, E. S., and Bain, E. C., " The Aging of Steel," Trans. Am. Soc. for Metals, **23**, 1047 (1935)

15. Merica, P. D., " Precipitation Hardening, Its Scope and Possibilities," Metal Progress, **27**, 31 (Jan. 1935) — and — **27**, 46 (March 1935)

16. Merica, P. D., " The Age-hardening of Metals," Trans. Am. Inst. Mining and Met. Eng., **99**, 13 (1932)

17. Keller, F., and Wilcox, G. W., " Polishing and Etching of Constituents of Aluminum Alloys," Metal Progress, **23**, 44 (April 1933) — and — **23**, 38 (May 1933)

18. Dix, E. H. Jr., and Keller, F., " Etching Aluminum For Metallographic Examination," Metals Handbook, 1290, Am. Soc. for Metals, Cleveland (1939)

19. Howell, F. M., and Paul, D. A., " Properties of Wrought Aluminum Alloys at Elevated Temperatures," Metals and Alloys, **6**, 284 (1935)

20. Gann, J. A., and Brooks, M. E., " Founding Magnesium Alloys," Trans. Am. Foundrymen's Assoc., **43**, 591 (1935) — and — **44**, 33 (1936)

DIE CASTINGS

Die Castings Compared with Other Cast and Wrought Products. — Permanent mold castings of the light alloys, white metals, bronzes, and even cast iron are made by pouring the molten alloy into metal molds containing sand or metal cores.[1] The rapid chilling of the metal where it contacts the cool metallic surface of the mold improves the mechanical properties generally. Aluminum and zinc alloys and the bronzes are employed in permanent mold practice, each in a rather distinct field.

Die castings differ from permanent mold castings in that the molten metal is forced into metal molds under considerable pressure. A somewhat similar process used for brasses and bronzes and other alloys with relatively high melting points involves injection at very high pressures of semifluid or pasty metal into a metal mold.

Some of the advantages of the die casting process are:

(1) Extremely rapid production compared with other methods of casting.

(2) Uniformity of shape and maintenance of very close dimensional tolerances, thereby reducing or eliminating machining.

(3) Production of castings with very thin sections.

(4) Casting of intricate shapes which would otherwise be assembled from several parts.

(5) Accurate reproduction of lettering, ornamental designs, and other details.

(6) Possibility of inclusion of inserts such as studs, bushings, magnetic or electrical parts within the casting.

The following list of dimensional tolerances for zinc base (Zamak Alloy) and aluminum base (particularly A.S.T.M. Alloys IV, V, and VIII) die castings indicates the relatively close dimensions which are maintained in this type of casting. In addition to

the listed variations, an additional tolerance of +0.006″–0.000″ may be expected on all unlimited dimensions which extend into both die-halves and depend upon positive closing of the dies or cores.

Lengths or Diameters in Inches	Nominal Variations in Inches	
	Zinc Base	Aluminum Base
Up to 0.5	±0.002	±0.004
0.5 to 1	0.003	0.005
1 to 2	0.005	0.007
2 to 3	0.007	0.009
3 to 4	0.009	0.011
4 to 5	0.011	0.013
5 to 6	0.013	0.015
6 and over	0.014	0.016

Courtesy of J. R. Townsend (by private communication).

Because of their uniformity, dimensional accuracy, and strength, die castings are able to compete with small forgings for many applications. Thin walled die castings also compete very actively with stamped, punched, and formed products. Internal combustion engine carburetors illustrate the possible intricacy of die castings.

A disadvantage of die casting is the high cost of dies, which usually limits the practice to parts produced in large quantities. Large intricate or multiple dies may cost several thousand dollars to produce. The continued rapid expansion of the die casting industry is ample evidence that there are many applications for which the cost of the dies is offset by the reduction in other manufacturing charges through the elimination of jigs, fixtures, gages, raw material inspection, machining, punching dies, drawing dies, patterns, screws, bolts, and other fastening devices.

Die casting dies are made of plain carbon, tool, and alloy steels. While the life of dies for zinc, lead, and tin alloys is usually limited only by mechanical wear, reduction of life from other causes becomes increasingly serious as the melting point of the alloy to be cast increases. Dies for aluminum die castings are somewhat shorter lived than dies for similar castings made from zinc alloys, yet they will produce several hundred thousand castings. In the case of the higher melting copper alloys, die life becomes a much

more serious problem. In any case die life is dependent on the shape of the part and the skill of the die designer with respect to gating and venting. The design of die casting dies is really a problem in hydraulics and careful attention must be paid to the characteristics of liquid metal flow.

The maximum weights and dimensions of die castings now in production are limited by the size of existing die casting equipment. Among the largest die castings being made at present are washing machine parts and automobile radiator grills. Some of these parts weigh from 10 to 15 pounds and their maximum dimension is as much as 30 inches. The ultimate size limits will probably be those imposed by the ingenuity of the die designer. The casting machines are being built in larger sizes and the operating pressures are being increased, thus extending the upper size and weight limits.

The Die Casting Process. — In contrast with sand casting practice in which the molding and pouring are often done by hand, die casting is strictly a machine operation. Several casting systems have been developed. Of these, the plunger type is most commonly used for the zinc alloys. A cylinder with a close fitting piston is submerged in molten metal. The piston, which may be actuated by air, hydraulic, or mechanical means, forces molten metal through a tube into a fixed section of the die. The movable portion of the die is held closed, often by hydraulic means, against the pressure within the mold. In machines of this general type, metal pressure up to about 3000 lb. per sq. in. is feasible. These machines are well adapted to automatic and high speed operation.

In another common type of machine, used largely for the aluminum alloys, the piston is replaced by direct air pressure on the molten metal. In these machines the pressure is limited to about 700 lb. per sq. in.

Much higher pressures are available in a machine of the plunger type in which the charge is melted apart from the injector and introduced into a relatively cold cylinder. A thin crust of solid metal forms on the walls of the pressure chamber and seals the piston during the working stroke. Aluminum and magnesium alloys, brasses, and many other copper alloys may be die cast by this method.[2]

In all of the above mentioned types of die casting machines, the die temperature is controlled by means of internal water passages. Because of the rapid freezing and the fine detail and thin sections which must be filled out, high operating pressures are needed. The rigidity of the metal dies confines the casting so that natural shrinkage is prevented in certain sections. It is necessary, therefore, to remove castings from the die as soon as possible after solidification in order to avoid the possibility of cracking caused by high internal stresses. The need for alloys free from hot-shortness is apparent.

Further details in die casting design, testing, finishing, and application will be found in two books on the subject [3, 4] as well as in other references for this chapter.

Desirable Characteristics of Die Casting Alloys. — Alloys to be used for die casting should have the following general characteristics. Naturally no one alloy excels in all these requirements.

(1) Low melting point — to reduce erosion and cracking of the molds.
(2) Low shrinkage.
(3) High fluidity — to facilitate making intricate or thin sectioned castings.
(4) Freedom from hot-shortness — to prevent cracking in the mold or upon ejection.
(5) Dimensional stability — to avoid growth or warping upon aging at operating temperatures.
(6) Good surface appearance and corrosion resistance.
(7) Good electroplating and organic finishing characteristics. (Not a requirement if the original surface is suitable for the application.)
(8) Adequate mechanical properties for the given application.
(9) Low cost.

Zinc Alloy Die Castings. — Die castings are produced in zinc, aluminum, magnesium, copper, tin, and lead alloys. The zinc alloys have many qualifications which justify their extensive use. They are low in cost, are readily cast at comparatively low temperatures, and have good mechanical properties. Their corrosion resistance and appearance may be improved by several finishing treatments including a short dip in alkaline chromate or dichromate

TABLE XIV. ZINC DIE CASTING ALLOYS

| | S.A.E. No. | 921 | 903 | 925 |
	A.S.T.M. No.	XXI	XXIII	XXV
Aluminum, %		4.1	4.1	4.1
Copper, %		2.7	—	1.0
Magnesium, %		0.03	0.04	0.03
Zinc, % (Special High Grade, 99.99% pure)		Remainder	Remainder	Remainder
Tensile strength, lb./sq. in.				
As cast		47 900	40 300	45 400
10 day steam, 95° C.		45 100	33 200	35 700
7 year indoor aging		48 700	32 900	37 400
Elongation, % in 2 in.				
As cast		5.1	4.7	4.9
10 day steam, 95° C.		1.0	5.5	1.8
7 year indoor aging		3.1	8.1	5.6
Impact strength, ft.-lb.				
As cast		19	20	18
10 day steam, 95° C.		1	22	11
7 year indoor aging		2	17	18
Dimension change, inches in 6 in.				
10 day steam, 95° C.		0.0128	0.0038	0.0033
7 year indoor aging		0.0091	0.0011	0.0007

Courtesy of W. W. Broughton, Metal Progress.[6] (Additional data are given in the original table.)

solutions and by the application of organic and electroplated finishes.[5] Baking temperatures for organic finishes should be below 380° F. to avoid blistering and possible dimensional changes.

The compositions and properties of modern zinc die casting alloys, as given by Broughton [6], are listed in Table XIV. Details of specifications, machining, and finishing are available in " Metals Handbook." [7]

In order to prevent the occurrence of intergranular corrosion, a condition which led to failure of many of the early zinc base die casting alloys, it is necessary to use high purity zinc and to keep the impurities lead, cadmium, and tin very low in the compositions given in Table XIV. The small addition of magnesium and the copper addition are necessary to obtain resistance to intergranular

corrosion beyond that available in the alloy without these elements. The effects of various minor elements have been summarized by Wright.[8] The steam test, as reported in Table XIV, has been found to be very dependable in detecting susceptibility of contaminated or improperly formulated alloys to intergranular corrosion.

Although absolute dimensional stability has not been attained, the small dimensional changes of the present alloys are of no importance for most applications. It has been found that the changes which occur accompany age hardening effects based on the limited solubility of aluminum and copper in zinc.[9] Aging treatments at temperatures between 70° and 100° C. are sometimes applied before machining when exceptionally close dimensional tolerances must be maintained.

The copper-free alloy of Table XIV is not the strongest but has the best ductility and toughness and the least susceptibility to aging effects. By storing the castings for five weeks or heating to 100° C. for six hours before machining, the changes in dimensions of this alloy after machining are less than 0.0003 inch per inch.[6]

Zinc alloys are subject to brittleness at low temperatures, whereas the other die cast alloys are not subject to this trouble.

Typical applications of die cast zinc alloys include automobile hardware, fuel pumps, carburetors, speedometers, machine parts, tools, locks, business machines, clocks, and novelties.

Light Alloy Die Castings. — The advantages of die casting are available in both aluminum and magnesium alloys. In addition to their light weight the aluminum alloys have good corrosion resistance, have the ability to take and hold a high polish, are stable at reasonably high temperatures, are not very sensitive to the effects of minor impurities, and have high thermal and electrical conductivity. They are not subject to low temperature brittleness though their impact strength at normal temperatures is not high. Because of their higher melting points they are somewhat more difficult to cast than zinc base and other die casting alloys having lower melting points.

Finishing methods for the aluminum die casting alloys include buffing to a high polish, production of a white frosted finish by an acid dip treatment, anodic oxidation, and the application of numer-

ous organic finishes including colored dyes on the oxidized surface.[5]

Aluminum-silicon alloys are suitable for die castings because they possess good corrosion resistance, freedom from hot-shortness, high fluidity, and other favorable characteristics.[10] A.S.T.M. Alloy V is one of the best all around casting alloys, however its anodic oxide coating is dark and has relatively low resistance to abrasion.

TABLE XV.   ALUMINUM DIE CASTING ALLOYS

A.S.T.M. No.	Composition * Per Cent			Tensile Strength lb./sq. in.	Elongation in 2 in., %	Charpy Impact ** ft.-lb.
	Si	Cu	Ni			
IV	5	—	—	29 000	3.5	4.5
V	12	—	—	33 000	1.5	2.0
VI	3	2	—	30 000	3.5	5.0
VII	5	4	—	32 000	2.0	2.5
VIII	—	1.5	2.25	29 000	4.0	4.5
IX	1.75	4	4	31 000	1.5	2.0
XI	8	2	—	32 000	1.7	3.0
XII	2	7	—	34 000	1.6	3.2

* Balance aluminum.
** Unnotched square specimens $\frac{1}{4}$ x $\frac{1}{4}$ inch.

From "Tentative Specifications for Aluminum-Base Alloy Die Castings — A.S.T.M. Designation: B85–37T," Proc. Am. Soc. for Testing Mats., 37, 705 (1937).

Note: The mechanical test data are *not* a part of the specifications but represent the properties that may be expected of unmachined tension and impact specimens when cast in a die in regular routine operation.

The aluminum alloys containing copper are cheaper, higher in yield strength and hardness, but lower in corrosion resistance than the aluminum-silicon types. High copper increases fluidity and reduces shrinkage but promotes hot-shortness. A.S.T.M. Alloy XII has been used extensively for low cost, general purpose castings. The copper containing alloys are easier machining, and have a lighter color than the silicon types after the anodic oxidation treatment, making them suitable for colored finishes except in the light shades.

The alloys with both copper and silicon present tend to combine the characteristics imparted by each, but corrosion resistance is inferior to the aluminum-silicon types.

The addition of nickel in Alloy IX makes this alloy whiter in color than the higher silicon types. A.S.T.M. Alloy VIII, also containing nickel, develops a very light colored, hard, and abrasion resistant oxide coating in the anodic oxidation treatment. However, it has relatively poor casting characteristics, hence it is generally used for small and simple castings.[5]

Some typical applications of the aluminum die casting alloys are household appliances, gas engine parts, automobile brake shoes, portable tools, and electrical equipment including motor frames and radio parts.

The principal magnesium base die casting alloys are the 10% aluminum alloy of Table XII, page 87, and a modified form of the same alloy containing 0.5 to 1.0% silicon to improve the fluidity. The tensile strength of the latter alloy is about 28,000 lb. per sq. in. and the elongation 1.0%. Magnesium alloy die castings are used for small parts and housings for vacuum cleaners, typewriters, tools and other portable equipment, and for aircraft parts. The application of these ultra-light castings will undoubtedly be greatly increased as their properties become more familiar to designing engineers.

Other Die Casting Alloys. — The first successful die castings were tin and lead base alloys which are easy to cast because of their low melting points. Tin base automotive engine bearings were formerly die cast, but the development of new bearing alloys and the use of centrifugal casting methods has decreased the use of die cast bearings. Tin base die castings are also used for certain corrosion resisting purposes. These include pewter and other special alloys for use with foodstuffs. The tin alloys are expensive, and, aside from their corrosion resistance, have properties inferior to those of the zinc base types.

Lead base die castings are used for special corrosion resisting articles as well as for X-ray equipment, toys, and cheap bearing metals. Lead base alloys containing tin and antimony are die cast as type metal in linotype and other printing machines.[11]

When high strength or special corrosion resistance, wear resistance, or electrical conductivity are desired, the use of copper base die castings may be warranted. The high melting points of the cop-

per alloys introduce difficult die making and casting problems. Although much progress has been made using very high pressure casting machines [2], the development of more suitable die steels would greatly extend the use of brass and bronze die castings for high strength parts for automotive, electrical, ordnance, marine, and other purposes. Brasses, bronzes, aluminum bronzes, and nickel silvers, to be discussed in Chapter VI, are among the alloys which are successfully die cast; however, the relatively low melting brass containing 60% copper-40% zinc is the most economical to cast in the dies now available.

REFERENCES

1. Marius, H., "Permanent Mold Foundry Practice for Bronze Castings," Trans. Am. Foundrymen's Assoc., **39**, 393 (1931)
2. Fox, J. C., "Brass Die Castings" (Appendix to the Report of Committee B-6), Proc. Am. Soc. for Testing Mats., **36**, Part I, 215 (1936)
3. Chase, H., "Die Castings," John Wiley and Sons, New York (1934)
4. Herb, C. O., "Die Casting," The Industrial Press, New York (1936)
5. Fox, J. C., "Finishing of Die Castings" (Appendix to the Report of Committee B-6), Proc. Am. Soc. for Testing Mats., **36**, Part I, 193 (1936)
6. Broughton, W. W., "Zinc Alloy Die Castings," Metal Progress, **33**, 381 (1938)
7. Anderson, E. A., and Stern, M., "Manufacture and Properties of Zinc Die Castings," Metals Handbook, 1770, Am. Soc. for Metals, Cleveland (1939)
8. Wright, T. A., "The Role of the Spectrograph and of Minor Elements in Die Castings," Trans. Am. Inst. Mining and Met. Eng., **117**, 329 (1935)
9. Fuller, M. L., and Wilcox, R. L., "Phase Changes During Aging of Zinc-Alloy Die Castings, II—Changes in the Solid Solution of Aluminum in Zinc and Their Relation to Dimensional Changes," Trans. Am. Inst. Mining and Met. Eng., **122**, 231 (1936)
10. Dix, E. H. Jr., and Bowman, J. J., "Effect of Composition on Mechanical Properties and Corrosion Resistance of Some Aluminum-Alloy Die Castings," Trans. Am. Inst. Mining and Met. Eng., **117**, 357 (1935)
11. Gonser, B. W., and Epstein, S., "Metals and Alloys in the Printing Industry," Metals and Alloys, **8**, 3 (1937)

COPPER AND ITS ALLOYS

It has been shown that the modern light alloys of aluminum and magnesium are now available in a wide range of compositions. The copper alloys, many of which have been of service to mankind for thousands of years, are being produced in an even greater range of compositions. In addition to zinc and tin, which are the principal alloying elements in ordinary brasses and bronzes, respectively, many other elements including nickel, beryllium, lead, aluminum, gold, and silicon form useful alloys with copper. The nominal compositions and properties of a large number of copper alloys are included in a comprehensive tabulation of proprietary engineering alloys compiled by Woldman and Dornblatt.[1]

A majority of the common alloying elements are soluble in appreciable amounts in solid copper. The resulting solid solutions have many valuable physical and chemical properties. Unfortunately, the electrical conductivity is drastically lowered by these elements, as little as 1 % of aluminum, silicon, iron, manganese, or phosphorus reducing the conductivity to less than one-half the normal value for copper.[2] Of the elements which are essentially insoluble in copper, lead is the most useful in copper alloys while bismuth is most objectionable because of its embrittling effect.

Copper for Electrical Purposes. — In order to produce high conductivity copper suitable for electrical purposes it is necessary to use very effective refining treatments to remove the impurities. Blister copper smelted from certain high grade Lake Michigan ores may be refined sufficiently by the fire-refining process; however, the major portion of the electrical copper is also refined by electrolysis which removes silver, gold, arsenic, antimony, and other impurities, some of which are removed only partially or not at all by fire-refining. The pure copper cathodes produced by electrolysis are melted, fire-refined, and cast into units suitable for processing into wire, rod, sheet, or tubing.

The fire-refining treatment is an oxidation process in which air

is introduced into the molten metal to oxidize or volatilize zinc, lead, iron, bismuth, and certain other impurities. Copper is also partially oxidized, therefore it is necessary to deoxidize the bath before casting. The deoxidizing agents used and the extent to which the metal is deoxidized determine, to a large extent, the type of copper obtained.

Tough pitch copper is made by deoxidizing with reducing gases introduced into the bath by stirring with green wood poles until the oxygen content is reduced to between 0.03% and 0.07%. The small residual oxygen content is necessary, in this process, to insure sound castings of high ductility from which the name " tough pitch " is derived.

The oxygen present in the solidified copper occurs as the oxide, Cu_2O. If approximately 3.5% Cu_2O were present (in insufficiently deoxidized copper) the melting point would reach a minimum of 1065° C., and the entire microstructure would become dotted with fine Cu_2O particles. This composition is the eutectic alloy of the Cu-Cu_2O system.[3] The cast structure of Fig. 9, page 12, contains only a small proportion of the $(Cu + Cu_2O)$ eutectic surrounding primary copper dendrites. The microstructures of wire and other products made from tough pitch copper ordinarily contain an even smaller number of well distributed Cu_2O particles. Excessive amounts of oxygen would lower the electrical conductivity and interfere with wire drawing operations, thus the final oxygen content must be carefully controlled in the fire-refining treatment.

Oxygen-Free Copper. — Elements such as phosphorus, cadmium, silicon, boron, barium, calcium, and lithium may be added to molten copper to remove the last remaining oxygen.[4] Phosphorus has been used extensively for this purpose, producing electrolytic phosphorized copper. If the additions are in excess of the requirements for deoxidation purposes, the electrical conductivity will be lowered; nevertheless, copper for non-electrical purposes may be produced in this way.

Carefully deoxidized copper produced by a special process is now commercially available under the name of oxygen-free high-conductivity (OFHC) copper.[5] It contains 99.98% copper and less than 0.005% of any one impurity. Its electrical and mechanical proper-

ties are essentially the same as those of tough-pitch copper, excepting that cold drawing and forming operations are more readily carried out because of its improved plasticity.

When tough-pitch copper is annealed in reducing gases containing hydrogen, hydrocarbons, or carbon monoxide, these gases penetrate the metal, reduce the oxides present, and form water vapor or carbon dioxide which remains entrapped, causing embrittlement or even disintegration at grain boundaries.[6] Obviously oxygen-free coppers are not subject to this destructive action in process annealing, thereby extending their availability to products requiring several cycles of cold deformation and annealing, including seamless tubing for water systems and other purposes. Bright annealing practice for ordinary and oxygen-free coppers is discussed by Hansen and Brace.[7]

Effect of Impurities in Copper. — The impurities in commercial coppers may be divided into two classes. Arsenic and phosphorus are representative of those which dissolve in copper to form solid solutions of low electrical conductivity. Although 0.1% phosphorus reduces the conductivity to about one-half that of pure copper, it is an excellent deoxidizer and improves the cold working properties. Copper containing about 0.02% phosphorus is often used for pipe and tubing. Even smaller amounts of residual phosphorus are present in deoxidized coppers for electrical applications. Copper containing 0.25 to 0.50% arsenic is used in place of steel for locomotive fire-box plates (especially in European countries).

Bismuth, sulphur, and lead are representative of the elements which are insoluble in solid copper. The electrical conductivity is only moderately affected by small amounts of these elements; in fact, many properties such as conductivity and specific gravity vary in a linear manner between the properties of two pure metals which form a mechanical mixture alloy. In such cases the effect of small amounts of impurities is very slight. On the other hand, when the atomic crystal lattice of a metal is disturbed by the presence of foreign elements in solid solution or through intermetallic compound formation, the resistance to the flow or transfer of electrical charges is increased. Plastic deformation also increases electrical resistance because of the resulting distortion of the crystal structure.

The type of crystal lattice bears a relationship to conductivity, the F.C.C. metals such as silver, copper, gold, and aluminum being the best conductors.

Although bismuth has relatively little effect upon conductivity, it is considered to be the most objectionable of the mixture-forming impurities because it forms a brittle film or network around the solid copper crystals, interfering with both hot and cold working when present in amounts over 0.01%. (See page 29.)

Copper Brazing. — Joining of assembled parts by a method known as copper brazing is an interesting application of pure copper. This form of welding is applicable to steel and other high melting alloys. The parts are rigidly assembled using screw joints, press fits, clamps, or other devices to insure close contact. Copper is applied at the joint or joints in the form of wire, sheet, paste, or powder. The assembly is then heated slightly over the melting point of copper in a reducing atmosphere whose principal constituent is hydrogen. The copper flows into the joint by capillary attraction, giving a gas tight weld which will withstand reasonably high pressures.[8] Copper brazed joints are used for such varied applications as refrigerator units, steel golf shafts, and tungsten carbide inserts for cutting tools.

Costs of Nonferrous Metals and Alloys. — The basic prices of many of the metals and alloys of importance in the brass and bronze industry follow. These prices are from " American Metal Market " for Jan. 4, 1939.

<div align="center">New Metals and Ingots</div>

Copper (electrolytic)	11.12¢/lb.
Zinc (brass special)	4.60 ”
Tin (Straits)	46.40 ”
Lead (soft Missouri)	4.70 ”
Aluminum (virgin ingot — 99+%)	20.00 ”
Nickel (ingot or shot)	36.00 ”
Monel metal (shot)	28.00 ”
Yellow brass ingot metal (No. 1)	8.87 ”
85–5–5–5 ingot metal	11.00 ”
80–10–10 ingot metal	12.87 ”
88–10–2 ingot metal	14.50 ”
Beryllium copper (4% Be), per lb. contained beryllium	$23.00

Soft wire .... 14.87¢/lb.
Drawn rod .... 16.87 ”
Hot rolled sheet .... 19.37 ”
Seamless tubes .... 19.87 ”

WROUGHT YELLOW BRASS

Sheet .... 17.31¢/lb.
Wire .... 17.56 ”
Rod .... 17.75 ”
Rod, free cutting .... 12.62 ”
Seamless tubes .... 20.06 ”

RODS

Hot rolled copper .... 15.87¢/lb.
Yellow brass .... 17.75 ”
Red brass (15% Zn) ... . 18.81 ”
Commercial bronze (10% Zn) .... 19.62 ”
Commercial bronze (5% Zn) .... 19.93 ”
Leaded brass .... 19.75 ”
Muntz metal .... 16.62 ”
Tobin bronze .... 19.37 ”
Manganese bronze .... 20.12 ”
Everdur, hot rolled (Grade A) .... 21.62 ”
Phosphor bronze (5% Sn) .... 34.00 ”
Nickel-silver (18% Ni) .... 34.75 ”
Monel, hot rolled .... 35.00 ”
Nickel, hot rolled .... 45.00 ”
Beryllium copper, per lb. .... $ 1.30

Copper-Lead Alloys. — About 70% of the copper produced is used in the form of wire, rod, sheet, strip, and tubing, the remainder being used for alloy making, including ordinary brass and many other compositions of technological importance.

For all practical purposes copper-lead alloys are of the mechanical mixture type. (The solid solubilities are: 0.06% Cu in Pb, 0.02% Pb in Cu.) The two distinct solid phases, copper (light) and lead (dark), are clearly visible in Fig. 47, the structure of a bearing alloy containing about 30% lead. Despite the apparent simplicity of their microstructures, the production of leaded-copper castings free from segregation and other defects is a difficult task requiring the use of special melting and casting practice, and, in most cases, the addition of small amounts of a third element or

elements to the composition.   One of the reasons for this tendency
to segregate is the fact that certain compositions are only partially
soluble in the liquid state.   Alloys containing more than 38% lead
form two liquid layers of different compositions just before the
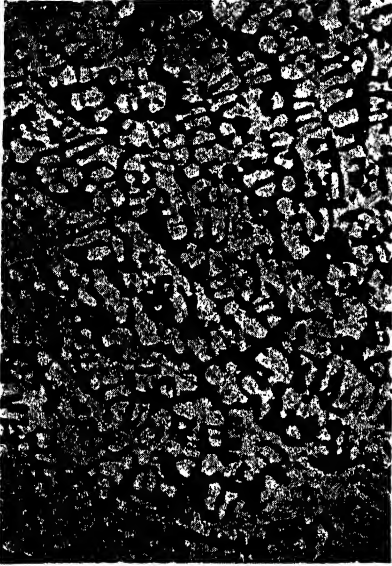
FIG. 47. — BEARING ALLOY CON-
TAINING 70% COPPER-30% LEAD.
(Unetched.)   X100. (*Courtesy of R.
L. Heath, Allison Engineering Com-
pany.*)

copper-rich phase solidifies.   Ade-
quate stirring of the melt and rapid
cooling of the casting is necessary
to maintain a uniform distribution
of copper and lead.   Additions of
nickel and silver are helpful in im-
proving the liquid solubility.   The
wide range of temperatures be-
tween the freezing of the copper
crystals at 954° and the lead crys-
tals at 326° C. gives further op-
portunity for mechanical segrega-
tion.   Equilibrium diagrams and
other pertinent data are given by
Jenkins [9] and Esborg.[10]

Fetz [11] has described methods
of avoiding segregation in leaded-
copper bearings by using com-
pressed metal powders sintered in
a hydrogen atmosphere. (See also
page 133.)

Copper containing 30 to 40%
lead has long been used for heavy duty bearings for railroad and
rolling mill service.   More recently the improved methods for pro-
ducing such alloys have made leaded-copper bearings available for
aircraft and automotive engines.   They are used successfully for
applications in which the loads or temperatures are excessive for or-
dinary white metal bearings.   The melting point of the lead in these
alloys controls their strength at elevated temperatures, which is, in
general, superior to that of the white metal bearing alloys.   High
thermal conductivity and very good frictional characteristics are
further advantages of leaded-copper bearing alloys.   (See also
page 131.)

Tin or nickel up to 10% is sometimes added to strengthen and toughen the copper matrix by solid solution formation. A typical composition of the modified type is 70% Cu, 27% Pb, 3% Sn.

Brass and Bronze Alloy Systems. — The equilibrium diagram for the copper-zinc system, Fig. 48, is typical of the complex com-
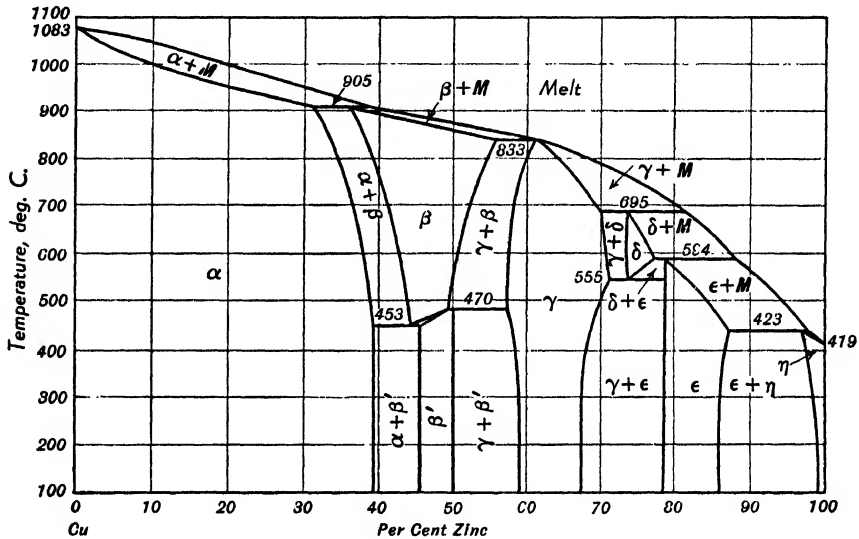
FIG. 48. — THE COPPER-ZINC EQUILIBRIUM DIAGRAM. (*Courtesy of L. R. Van Wert, An Introduction to Physical Metallurgy, McGraw-Hill Book Company, 1936.*)

binations of solid solutions and compounds which occur in the Cu-Zn, Cu-Sn, Cu-Al, Cu-Be, and Cu-Si systems. (See also Figs. 68, 76, and 78.) Fortunately, the useful alloys of these systems generally occur in the high copper solid solution regions. In each case the alpha solid solution has the face-centered cubic crystal structure of copper, modified by the presence of the second element in solid solution. All of the alpha solutions are characterized by excellent cold working properties; in fact, within certain composition ranges the alpha solid solution alloys are even more ductile than copper itself. The second and third phases, from the copper side of the diagrams, have crystal structures which are less plastic; beta brass and delta aluminum bronze, for example, being quite brittle.

Notwithstanding important commercial developments in a wide

variety of copper alloy systems, the zinc brasses remain the most important copper alloys from the tonnage standpoint. Ordinary alpha brasses have a desirable combination of properties, including excellent cold forming characteristics, good corrosion resistance for ordinary applications, and low cost. In the form of sheet, rod, and wire for cold fabrication they have innumerable industrial uses.

The nominal compositions of important brasses, based mainly on A.S.T.M. specifications, are listed in Table XVI.

TABLE XVI. NOMINAL COMPOSITIONS OF BRASS

NAME	PER CENT COMPOSITION				
	Cu	Zn	Pb	Fe	Sn
Commercial bronze	90	10	—	—	—
Red brass	85	15	0.07 max.	0.07 max.	0.15 max.
Low brass	80	20	—	—	—
Malleable brass	75	25	—	—	—
Admiralty metal	70	29	0.075 max.	0.06 max.	0.90–1.20
Cartridge brass	70	30	0.07 max.	0.05 max.	—
High brass	68	32	0.07 max.	0.04 max.	—
Common high brass	66	34	0.35 max.	0.06 max.	—
Free-cutting brass	61.5	35.5	2.5–3.75	0.15 max.	—
Brass forging rod	60	38	1.5–2.5	0.15 max.	—
Muntz metal					
(tube grade)	60	40	0.30 max.	0.07 max.	—
Naval brass	60	39	0.30 max.	0.10 max.	0.5–1.0

Based mainly on A.S.T.M. Standards.

Commercial Bronzes. — The nomenclature of the brass industry, although adequate for its purpose, is often misleading to the student or layman. The term " bronze," unless qualified by a second term, refers to a copper-tin alloy. Aluminum, silicon, and beryllium bronzes, and the commercial bronzes about to be discussed do not contain tin.

The 5 to 10% zinc alloys range from copper-red to bronze in color and are used for wrought articles requiring better corrosion resistance and appearance than the higher zinc brasses. The 10% zinc alloy is known as commercial bronze. These alloys may be hot worked from the ingot but are usually finished by cold rolling or drawing. Unlike high zinc brasses they are not subject to certain

types of corrosion known as season cracking and dezincification. Commercial bronze has been used for automobile radiators, screens, electrical fittings, and articles to be given a baked enamel finish.

Red Brass. — The composition for red brass given in Table XVI is that of a wrought alloy used mainly for brass pipes. As in the case of commercial bronze, it is not subject to failure by local dezincification.

The term red brass is also applied to a group of casting alloys having low zinc contents with additions of lead and tin. They are more resistant to corrosive waters, steam, and atmospheric exposure than the higher zinc brasses, and are widely used as castings for water, steam, gas, and electrical fittings and valves. The well known 85% Cu, 5% Zn, 5% Pb, 5% Sn alloy, included on the above price list as 85–5–5–5, is typical of these casting alloys.[12] Using the same notation, other common red brasses may be expressed as: (84–10–3–3), (80–10–7–3), (83–3–7–7), and (88–4–2–6). The latter composition, also known as " Steam Metal," has been used extensively for valve bodies.

The mechanical properties of red brass castings are not high, usually 25–35,000 lb. per sq. in. tensile strength, 15–35% elongation, and 40–60 Brinell hardness. Tin bronzes are used for similar applications where service stresses are higher or corrosive conditions are more severe. However, the red brasses are easier to cast and machine and are cheaper, consequently they are usually preferred, especially for valves and fittings and many allied parts to be finished by electroplating.

Low Brass. — Low brass is a wrought product having high ductility and a brilliant golden color. It is readily worked into cheap jewelry, hardware, tubing, and similar articles requiring an attractive appearance or good corrosion resistance. The malleability is high in this composition range, reaching a maximum in the 25% zinc alloy, which may be hammered into foil resembling gold.

Admiralty Metal. — This composition is used as tubing for marine condensers, preheaters, evaporators, and heat exchangers in contact with fresh and salt water, oil, steam, and other liquids at temperatures below 500° F. Its resistance to corrosion is superior to that of ordinary tin-free brasses for this type of service.[13]

Cartridge Brass. — This grade has the highest tensile ductility of the copper-zinc alloys. It must be very low in impurities such as lead and iron to withstand severe cupping or drawing operations.

High Brass. — High brass or yellow brass contains 31 to 35% zinc. The purest grade is a spinning and deep drawing brass having

Left. FIG. 49. — CAST BRASS CONTAINING ABOUT 33% ZINC. (All photomicrographs in this chapter were etched with an H_2O_2–NH_4OH solution unless otherwise stated.) Rockwell " F " 66. X25.

Right. FIG. 50. — STRUCTURE OF CASTING AFTER HEATING FOR 2 HOURS AT 800° C. (Etched.) Rockwell " F " 59. X75.

a zinc content which is usually at the lower end of the above range. Common high brass is made from a less pure spelter and usually contains 33 to 35% zinc. It is used where forming and physical requirements are somewhat less severe.

The remaining brasses of Table XVI, to be described later, do not have simple alpha solid solution structures.

The mechanical properties and microstructures of the alpha brasses have been discussed at length by Bassett and Davis [14], Pratt [15, 16], and others.

Solidification and Structure of Cast Alpha Brass. — The alpha

brasses will be used to extend the study of solid solution alloys begun in Chapter II. The wide application of the principles involved to ferrous, as well as nonferrous alloys, warrants their special consideration.

The cored dendritic structure of Fig. 49 is typical of many cast solid solution structures. The difference in shading from one dendrite to another is evidence of differences in their orientations. In general, the rate and manner of etching depends upon the orientation of the surface exposed; however, color effects from this source vary considerably with the alloy. Tin bronzes and very low carbon steels, for example, are quite uniform in color after ordinary etching procedures, while brasses are subject to marked variations in shading depending on the orientations of the crystals or dendrites.

The origin and nature of this heterogeneous cast structure is best studied by means of the equilibrium diagram of Fig. 48. When the 33% zinc alloy cools, the first solid to freeze contains approximately 23% zinc, consequently the central, light etching portions of the individual dendrites of Fig. 49 have a lower zinc content than the average composition of the alloy. As cooling proceeds the dendritic skeletons are built up with metal increasingly rich in zinc, but since little time is available for diffusion, the last material to solidify will have a much higher zinc content than 33% — probably high enough to produce a small amount of the beta phase. In this condition the alloy is not capable of withstanding the severe plastic deformation to which the wrought alloys are subjected. However, the first working and annealing operations will tend to homogenize the highly segregated structure and improve the ductility.

Homogenization by Annealing. — Upon heating the casting to 800° C. for two hours the marked change shown in Fig. 50 was effected. Although granulation of the dendritic structure is quite complete, the short black lines are regions of high zinc content, corresponding to the interdendritic or " filling-in " portions of Fig. 49. In this condition the ingot is somewhat softer and much more ductile than in the cast condition. Complete homogenization requires more time (or a higher temperature) but finally results in a granular structure free from dendritic markings. Such a structure
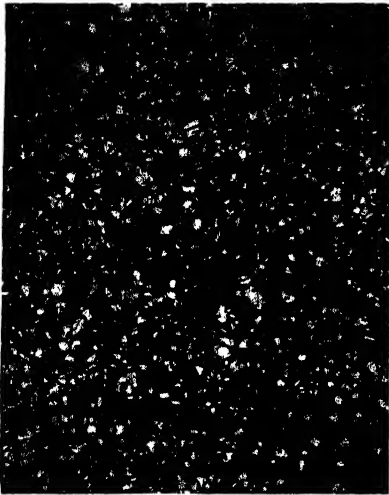
FIG. 51
(0.015 mm.)
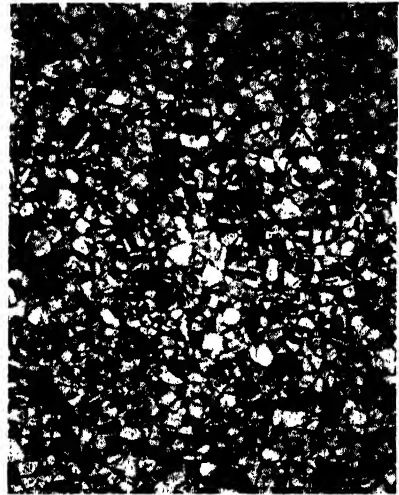
FIG. 52
(0.025 mm.)

FIG. 55
(0.065 mm.)

FIG. 56
(0.090 mm.)

FIGS. 51 TO 58. — GRAIN SIZE STANDARDS FOR THE ESTIMATION OF THE DIAMETER
SUCH AS BRASS, BRONZE, AND NICKEL-SILVER. (A.S.T.M. Standard E2–36.) Magnifi-
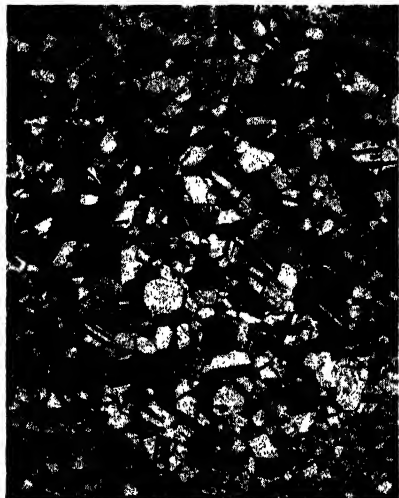
FIG. 53
(0.035 mm.)
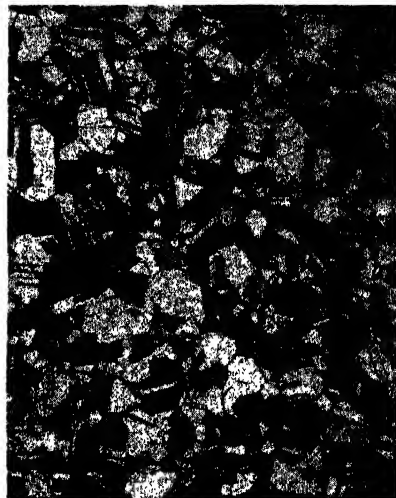
FIG. 54
(0.045 mm.)
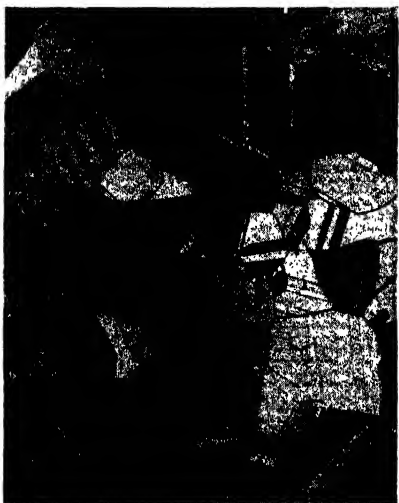
FIG. 57
(0.120 mm.)

FIG. 58
(0.200 mm.)

OF AVERAGE GRAIN OF ANNEALED MATERIALS, PARTICULARLY NONFERROUS ALLOYS
cation X75. (*Courtesy of Am. Soc. for Testing Mats.*)

has already been shown in Fig. 6. (Fig. 6 also shows slip lines produced by mechanical deformation after polishing.)

Homogenization by Working and Annealing. — The annealed casting of Fig. 50 was given a heavy reduction by cold rolling, followed by annealing at 500° C., whereupon a fine recrystallized grain structure was obtained. The average diameter of the grains was approximately 0.015 mm. and the structure was very similar in appearance to the A.S.T.M. grain size standard of Fig. 51. Further annealing at 800° C. gave a structure similar to Fig. 56.

Each grain of these cold worked and recrystallized structures is uniform in composition and chemically equivalent to every other grain. Homogenization by this method is usually very effective.

A.S.T.M. Standard E2–36. — The photomicrographs of Figs. 51 through 58 are from a series prepared by the American Society for Testing Materials for use in classifying grain sizes of recrystallized brasses, bronzes, nickel-silvers, and other homogeneous solid solution alloys. In each case the grain size is the actual diameter of the average grain.* (Two additional sizes, 0.010 mm. and 0.150 mm., are included in the A.S.T.M. Standard.)

Twinning. — The effect of surface orientation on the colors produced by the etching reagent is very marked in recrystallized brasses. The color banding which is prominent in the structures of Figs. 51 to 58 is evidence of twinning, a phenomenon associated with regional changes in orientation within individual crystals. Although the mechanism of twin formation [17] is rather complicated, the conditions necessary for twinning are quite clearly defined.

Annealing twins occur when certain metals and solid solution alloys are mechanically worked and then annealed or recrystallized. They are most common in face-centered cubic structures. The internal stresses set up during casting or quenching are sometimes sufficient to induce annealing twins upon subsequent heating. Aluminum and its alloys, although face-centered cubic, are exceptional in that they rarely form twinned crystals.

Mechanical twins often occur, especially in body-centered cubic and hexagonal metals, as a result of mechanical deformation.

* The grain size may be determined from photomicrographs at X100 by comparing with the standard X75 photomicrographs, then multiplying the apparent size by 0.75.

Twins of this origin, called Neumann bands, are narrower than most annealing twins, often appearing as straight lines similar to slip lines. (See pages 6 and 153.)

Effect of Cold Working on the Mechanical Properties. — Cold rolling and cold drawing are becoming increasingly important metallurgical processing operations for both ferrous and nonferrous alloys. The advantages of cold over hot reduction include improved surface conditions, higher dimensional accuracy, improved mechanical properties, and usually lower cost. Certain alloys, for example the high brasses, have relatively poor hot working characteristics and are nearly always worked cold, while Muntz metal, which is readily forged and hot worked, is generally hot reduced except for the finishing operations. However, the advantages of cold reduction also apply, to a large extent, to hot reduced and cold finished stock. The many desirable qualities of cold finished steel bars are reviewed in Chapter VII.

Plastic metals withstand a high percentage of reduction by cold working without rupture; nevertheless, the mechanical properties change progressively as a result of strain-hardening. The effect of cold drawing on the properties of a high brass is shown in Fig. 59.[16] After 90% reduction the tensile strength is increased to 140,000 lb. per sq. in. while the ductility is greatly reduced. It is important to note that the elongation and reduction of area curves refer to the results of tensile tests applied to material which has been cold drawn by the amounts shown on the abscissa. Further reduction in size over 90% would be difficult to carry out and would finally result in rupture. These results may be compared with the stress-strain curves for cold rolled iron on page 14.

The processing schedule may be so arranged that the desired combination of mechanical properties will be attained when the stock has been reduced to finish dimensions. For example, a tensile strength of 80,000 lb. per sq. in. with 12% elongation and 65% reduction in area may be obtained by a final cold reduction of 35%, starting with soft annealed stock.

As in the case of aluminum and other alloys, brasses are supplied in various " tempers " based on the amount of cold reduction after the final anneal. The temper of high brass sheet is determined by

the number of B. and S. gage numbers reduction in thickness, according to the following schedule from A.S.T.M. Standard B36–33.

TABLE XVII.   TEMPER DESIGNATIONS FOR BRASS SHEETS

Temper	Reduction in B. and S. Numbers	Tensile Strength lb./sq. in.	Rockwell " B " (samples over 0.040 inch thick)
¼ hard	1	46 000–56 000	30–60
½ hard	2	53 500–63 500	50–73
¾ hard	3	61 000–71 000	70–80
Hard	4	68 000–78 000	78–85
Extra hard	6	79 000–88 500	85–89
Spring	8	86 000–95 000	88–92
Extra spring	10	89 500–98 500	89–93

Conditions at Slip Planes and the Effect of Low Temperature Annealing on the Mechanical Properties of Cold Worked Metals. — Cold worked metal is under considerable internal stress, which may be undesirable from the standpoint of intergranular corrosion (see page 121) and dimensional permanence, consequently a low temperature stress-relieving annealing treatment may be necessary. The effect of annealing temperature on a cold reduced high brass is shown in Fig. 60.[16]   It is apparent that heating to temperatures below 200° C. has comparatively little effect on the tensile properties, but does somewhat increase the tensile strength (and hardness) and lower the ductility.   The proportional limit and yield strength (not shown in Fig. 60) are increased to a greater extent than the ultimate tensile strength, consequently cold rolled or cold drawn spring materials may be improved by a low temperature annealing treatment.   This change in properties upon heating cold worked metals and alloys has been quite generally observed, and in the case of iron or mild steel the strengthening effect is particularly marked.   Storing for relatively long periods at room temperature often produces a similar increase in proportional limit and other elastic properties.

Jeffries and Archer [18] have reviewed this subject in detail in " The Science of Metals."   The following excerpts from their discussion of " Conditions at Slip Planes " (page 202) are given
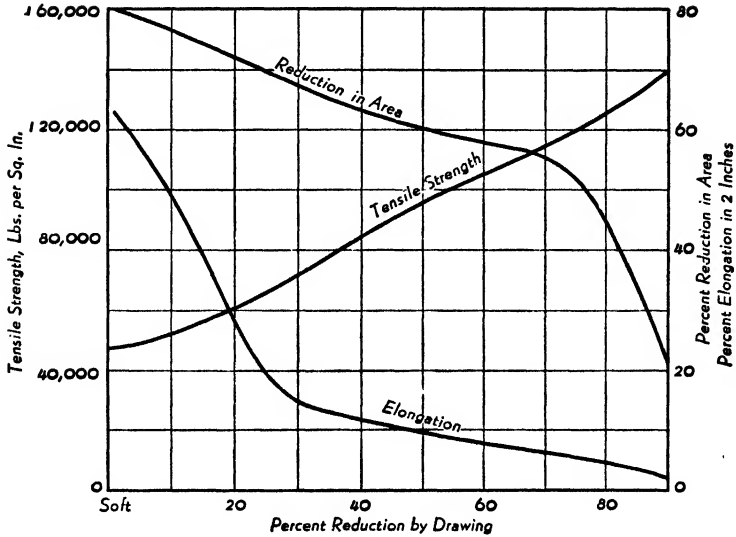
FIG. 59. — EFFECT OF COLD DRAWING ON THE PROPERTIES OF HIGH BRASS. (*Courtesy of R. S. Pratt, Metals Handbook.*[16])
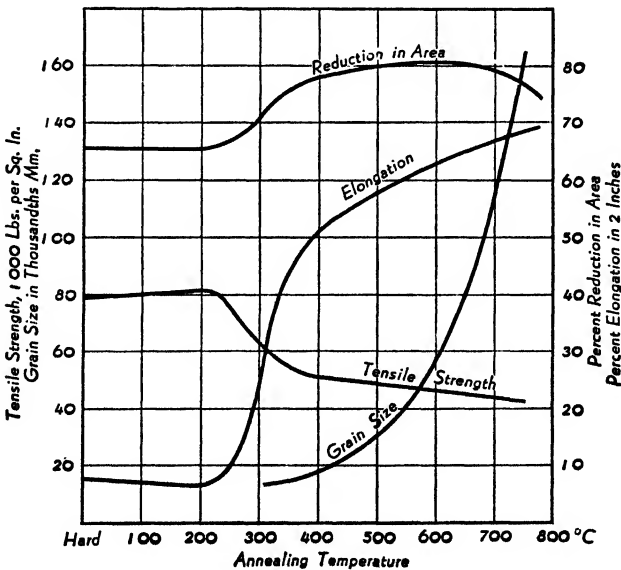
FIG. 60. — EFFECT OF ANNEALING TEMPERATURE ON THE PROPERTIES OF HIGH BRASS. (*Courtesy of R. S. Pratt, Metals Handbook.*[16])

as a review and extension of the general principles of plastic deformation (which were presented in Chapter I), and as an indication of a possible mechanism of strengthening cold worked metals by low temperature annealing.

"Immediately after slip begins, the slip plane strength is less than the crystal strength. . . .

"As slip continues, the slip plane strength increases to a value which may be greater than the crystal strength. . . .

"In a metal composed of an aggregate of grains, slip is stopped partly by the interference of adjacent grains and partly by the resistance on the slip plane itself. . . .

"Slip causes rupture of the atomic bonds on the slip plane, and immediately after motion has stopped there is only partial reestablishment of cohesion.

"When the registry of the displaced crystal fragments permits, cohesion is reestablished by the fragments joining into larger crystalline units. The slip plane then disappears as such, being replaced by a potential slip plane whose strength is equal to the crystal strength.

"As a rule, the crystal fragments do not register after deformation, and all degrees of disregistry occur. It has been shown by x-ray analysis that new orientations are created by plastic deformation. Consequently, there must be many, probably a large majority, of crystal fragments whose orientations do not permit them to unite except by the process of grain growth.

"Reestablishment of cohesion between crystal fragments of different orientations must be attended with various degrees of disorganization as regards the arrangement of the atoms at the slip plane.

"This metal of partly disorganized structure simulates an amorphous material in its mechanical properties. . . .

"Atomic rearrangement on slip planes takes place at temperatures much lower than are usually associated with recrystallization. During the spontaneous aging of overstrained iron at ordinary temperature, or the rapid recovery of elasticity at a blue heat, the slip plane strength increases so that small stresses no longer produce permanent deformations. This increase in slip plane strength,

or ' healing,' as it may be called, may consist in the growing together of fragments of sufficiently similar orientation, or in the establishment of cohesion at additional places on the planes between fragments that do not register. . . . All evidence is to the effect that the healing process involves an increase in the continuity of the metal."

The reader is referred to " The Science of Metals " for further elaboration of these topics and application of the slip interference theory to " blue brittleness " of iron and steel, mechanical properties at low temperatures, the nature of the material in the slip planes, and other related subjects.

Process Annealing of Cold Worked Metals. — Returning to the mechanical properties of annealed high brass, as reported in Fig. 60, marked changes in properties and grain size take place upon heating to higher temperatures above the recrystallization temperature range. Process annealing, carried out between 400° and 700° C. in the case of high brass, restores the plasticity, permitting further cold deformation by rolling, drawing, forming, or spinning of sheets; drawing of wire or rod; or cold heading of wire or rod in bolt and rivet making. In cases where high total reductions in size are required, it may be necessary to apply several cycles of cold working and annealing.

Cold Drawing, Forming, and Spinning. — Sheet metals are often fabricated by drawing or forming in dies or by spinning methods, all of which require a high degree of plasticity. The extremely soft, coarse grained structures obtained by annealing at high temperatures may be sufficiently plastic, but are likely to yield rough, coarse textured surfaces in the drawing or forming operation. Extreme softness may also cause buckling in flanges or other regions which are subjected to compressive stresses in the dies. For these reasons the finest grain size which can be fabricated successfully is usually desired. Ordinarily the temper and grain size requirements are determined for each job and then maintained within close limits by control of processing operations in the brass, aluminum, or steel mill.

Forming operations include a wide variety of fabricating procedures. A die forming set-up is sketched in Fig. 61.[19] If a high

hold-down pressure is used, it is obvious that considerable stretching or drawing will be necessary to complete this operation. Corrugating of sheets between rolls is an example of a forming operation involving little general elongation.

Drawing of sheet metals takes place under somewhat different conditions in that the metal being worked is, to a large extent, confined between the two dies, and considerable general elongation is involved, as shown by Crampton in Fig. 62. Although surface
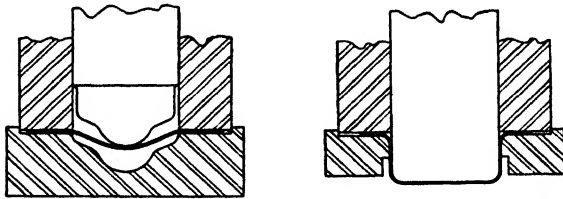
Left. FIG. 61. — DIE FORMING OPERATION. (*Courtesy of D. K. Crampton, Metal Progress.*[19])

Right. FIG. 62. — DRAWING OPERATION. (*Courtesy of D. K. Crampton, Metal Progress.*[19])

coarsening may occur at the rounded corners, the drawn sections are likely to be smoother than in the case of Fig. 61, where the metal is squeezed by the dies only at the end of the stroke.

Spinning is a lathe operation in which a flat disc of sheet metal is changed to a more or less cylindrical or cup shaped form by tool pressure applied while it is being rotated rapidly about its center.

The effect of thickness on the tensile properties of a deep drawing brass is illustrated by Crampton's data in Fig 63.[19] For a given grain size both the strength and ductility are reduced as the thickness decreases, and a drastic reduction in ductility occurs in the very coarse grain sizes. Although these results may possibly be accredited to changes in the number of grains across the thickness of the tensile test specimen, rather than to inherent differences in properties for a given grain size, the conditions in the dies are somewhat similar to those in the test specimen and greater difficulties are normally expected in drawing thinner sheets.

Season Cracking. — The atmospheric corrosion resistance of high zinc brasses is ordinarily satisfactory unless certain conditions

of internal stress are set up which promote disintegration of the metal by a process known as season cracking.[20]

Internal stresses produced in the fabrication of brass articles containing more than 20% zinc may sensitize the metal to selective corrosion at grain boundaries, which may lead to complete failure, with or without the application of additional stress in service. In
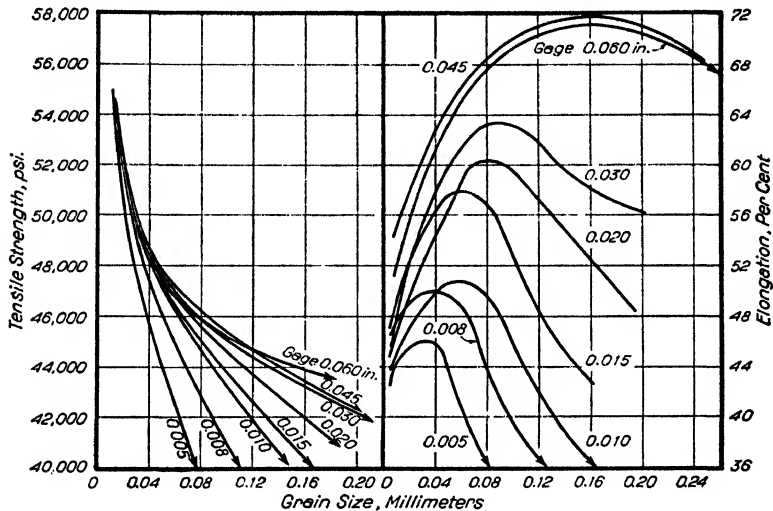
FIG. 63. — TENSILE PROPERTIES OF EYELET BRASS (68% CU-32% ZN) OF VARIOUS GAGES AND GRAIN SIZES. (*Courtesy of D. K. Crampton, Metal Progress.*[19])

severe cases failure occurs after a period of storage in ordinary atmospheres. Corrosive atmospheres and certain specific reagents, such as mercurous nitrate and ammonia salts, cause much more rapid disintegration and are commonly used for short time tests of susceptibility to season cracking.[21] It has been found possible in many cases to control the reductions and draws in such a way as to reduce the residual internal stresses and prevent season cracking. In other cases low temperature annealing treatments are employed to relieve the stresses without reducing the strength.

Brass containing 30 to 40% zinc happens to be most subject to this type of corrosion. However, many other commercial metals, from the light alloys to boiler steels, are more or less susceptible to intergranular stress-corrosion attack. The distinction between

stress-corrosion cracking, season cracking, and intercrystalline corrosion is largely a matter of the type and amount of stress to which the metal is subjected while under corrosive conditions.

Mitchell [22] describes the general types of corrosion to which copper alloys are subject.

Leaded Brasses. — Small amounts of lead, up to about 4%, are added to brasses to improve their machinability.[23, 24] Although cold forming properties are impaired, a high zinc brass containing 2% lead is suitable for hot forging and pressing. (See Table XVI, page 108.) As in the case of the copper-lead alloys the lead is essentially insoluble, therefore lead particles are distributed throughout the structure and tend to break the chips in high speed machining operations.

Muntz Metal. — When the zinc content of brasses exceeds about 39%, the beta phase becomes prominent. Between 39 and 45.5% zinc the wrought alloys are mixtures of the F.C.C. alpha solid solution and the B.C.C. beta solid solution. (See Fig. 48, page 107.) The beta phase may be present in cast alloys with lower zinc contents (as low as 30 to 35%), indicating departure from equilibrium conditions. Beta brass is much less plastic than alpha, and alloys which contain fairly large proportions of the beta phase must be worked while hot when heavy reductions by forging or rolling are necessary. The alloys containing 39 to 41% zinc, usually known as Muntz metal, may be readily extruded, rolled, or forged while hot, then finished by cold rolling or drawing.

While beta brass itself is inherently coarse grained and brittle and has but limited commercial application, it is possible to make fine grained alpha plus beta brasses having good mechanical properties. The normal cast structure of a 40% zinc alloy is shown in Fig. 64. The equilibrium diagram, Fig. 48, indicates that this alloy freezes as beta solid solution of somewhat variable composition, and transforms partially to alpha at lower temperatures. The transformation takes place rather sluggishly, therefore the cast alloy retains a larger proportion of the beta phase than equilibrium conditions would permit.

Close examination of Fig. 64 is warranted because it is typical of many cast alloys having duplex structures. A cross section of one,

Left. FIG. 64. — MUNTZ METAL (60% CU-40% ZN). AS CAST. (Etched.) 75 Brinell. X100. *Right.* FIG. 65. — MUNTZ METAL (60% CU-40% ZN). CAST — COLD HAMMERED. (Etched.) X100.

Left. FIG. 66. — MUNTZ METAL (60% CU-40% ZN). CAST — HEATED TO 825° C. AND COOLED VERY SLOWLY. (Etched lightly with H_2O_2–NH_4OH solution.) X100. *Right.* FIG. 67. — MUNTZ METAL (60% CU-40% ZN). CAST — HEATED TO 825° C. AND QUENCHED IN ICE WATER. (Etched.) 125 Brinell. X100.

and portions of four other original beta grains are included in the area shown. The white reed-like regions are the alpha phase which gradually separated from the beta upon cooling in the mold. In cast steels and other alloys we likewise find a coarse grain structure established by the elevated temperature phase, and a fine secondary structure made up of the products of transformation.

The effect of severe distortion by hammering is shown in Fig. 65.

The effect of heating to the beta field and cooling very slowly in the furnace is shown in Fig. 66. The alpha phase is coarsened (compared with Fig. 64) because the low cooling rate allows sufficient time for transformation and growth of the alpha crystals. Heating to 825° C. and quenching in cold water prevents the trans·formation to alpha except at grain boundaries as in Fig. 67. (Note that under these conditions the alpha phase rather than the beta phase is darkened by the etching reagent.) This is an example of hardening by heat treatment, since the hardness of the beta solid solution is approximately 125 Brinell compared with 75 Brinell for the original duplex structure. By reheating the quenched structure to 250° C., the supersaturated beta solid solution may be further hardened to about 230 Brinell by the precipitation of fine alpha particles.[25] However, Muntz metal is seldom heat treated in this way because of the brittleness of the product. If greater hardness is desired, the zinc content may be increased to give a greater proportion of the beta phase, or certain alloying elements may be added as described later.

Ordinary sand cast Muntz metal has mechanical properties comparable to those of mild steel; namely, 50–60,000 lb. per sq. in. tensile strength and 15 to 30% elongation, while the wrought alloys are even more ductile. Muntz metal has been rather widely used for water pipe and steam condenser tubing, but is not as corrosion resistant as other more expensive copper alloys now used, for example, for marine steam condensers. It has many other applications in both cast and wrought forms.

Modified Muntz metal alloys containing approximately 61% Cu, 1% Pb, 1% Sn, and 0.1% Al are successfully die cast in alloy steel molds.[26]

Tobin bronze, widely known and used as brazing rod, is a care-

fully processed Muntz metal type alloy containing 0.5 to 1% tin and approximately 39% zinc. It has higher mechanical properties than ordinary 40% zinc brass and is also used for pump parts, instruments, valves, and underwater machine parts for boats.

Manganese Bronze. — Manganese bronze is actually a high strength casting alloy based on the fundamental 60% Cu-40% Zn brass analysis. The composition range given in A.S.T.M. Standard B54–27, and data for three typical alloys reported by Ellis [27] are given in Table XVIII.

TABLE XVIII. TYPICAL MANGANESE BRONZES

	A	B	C	A.S.T.M. Standard B54–27
Cu, %	56.1	56.4	58.2	55–60
Mn, %	0.8	0.9	0.7	0–3.5
Al, %	0.4	1.7	0.8	0–1.5
Fe, %	1.0	1.1	1.0	0–2.0
Sn, %	—	—	0.25	0–1.5
Zn, %	Balance	Balance	Balance	42–38
Tensile strength, lb./sq. in.	78,000	89,200	76,700	65,000
Elongation in 2 in., %	35	21	26	25
Method of casting	Chill	Chill	Sand	Sand

According to Ellis, ". . . it seems logical to consider the high strength brasses as aluminum brasses to which small proportions of iron and manganese are added, the former for the purpose of increasing ductility without reducing the strength; the latter for the purpose of increasing the strength without reducing the ductility." [27] However, there is little agreement on the functions of the individual elements in these complex alloys. For example, aluminum has been recommended to improve the fluidity and other casting qualities by one authority, while another believes that it causes coarse crystallization and increased shrinkage. Further details of foundry practice, compositions, and properties are given by Ellis [27], Murphy [28], and Hensel.[29]

Manganese bronze is well known as an alloy for ship propellers.[30] Other applications for marine service, ordnance, machine parts, and electrical equipment depend upon its resistance to corro-

sion, good mechanical properties, non-magnetic properties, and comparatively low cost.

In addition to the brasses whose structures contain both the alpha and beta phases, alloys with zinc contents between 43 and 46% are sometimes cast. The microstructures of these copper-zinc alloys consist essentially of the beta phase, which increases the hardness at the expense of ductility. However, Murphy reports that beta manganese bronze castings have tensile strengths up to 112,000 lb. per sq. in. with 15 to 20% elongation.[28]

An alloy somewhat similar to manganese bronze but having lower zinc content and greater additions of the hardening elements, aluminum, manganese and iron, is used for cast trunnion bearings and miscellaneous heavily loaded bushings for ordnance purposes. Its minimum requirements are 100,000 lb. per sq. in. tensile strength and 15% elongation in 2 inches.[31]

Copper-Tin Bronzes. — True bronzes are copper-tin alloys. The first investigation of their equilibrium relationships, reported in 1903 by Heycock and Neville[32], introduced modern methods of investigation of alloy systems, including the use of cooling curves and photomicrographs.

The science of metallography came into existence only thirty-nine years before Heycock and Neville's work on bronzes. The first application of microscopic methods to the examination of metal structures was the study of meteoric irons in 1864 by the English scientist, Henry Clifton Sorby. The term "metallography" has since been applied to the entire science of development, fabrication, heat treatment, testing, and application of metals and alloys. However, the alternative expression "physical metallurgy" is often preferred, probably because "metallography" has become too closely associated with the purely microscopic aspects of the science.

The late Dr. Albert Sauveur, who was a contemporary of Heycock and Neville as well as a modern authority on metallurgy, is responsible for much of the progress made in the metallography of iron and steel. His task in 1893 was to apply the little used microscopic methods to practical steel making and treating. This was accomplished without the advantages of modern equipment, for even though it had been available the executives of that time were
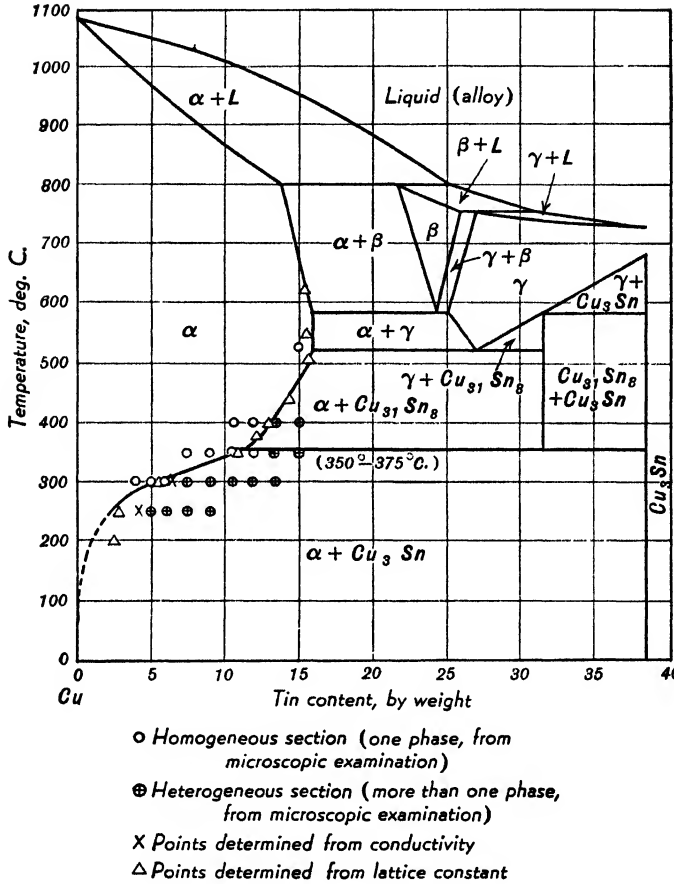
Fig. 68. — The Copper-Tin Equilibrium Diagram. (*According to C. Haase and F. Pawlek, Trans. Am. Inst. Mining and Met. Eng.*[33] — *See also Metals Handbook, p. 1364, 1939.*)

not inclined to provide facilities for such seemingly non-productive enterprises.

The complicated copper-tin equilibrium diagram, the copper-rich portion of which is reproduced in Fig. 68, has been the subject of much research since Heycock and Neville's pioneering investigation and several changes have been proposed in recent years. (Compare Fig. 68 with Dews.[34])

Phosphor Bronzes. — The wrought alpha bronzes, often called phosphor bronzes, contain between 1.5 and 10% tin. The amount

of phosphorus used may be just enough for deoxidizing purposes, or a phosphorus content of several tenths of one per cent may be left in the alloy, in which case it is even more effective than tin in increasing the hardness and strength, but it reduces the ductility. According to some classifications, phosphor bronzes are those containing more than 0.03% phosphorus, while those with lesser amounts are called phosphorus-treated alloys. Small amounts of phosphorus increase the fluidity and reduce gassing during solidification. Melting practice for wrought phosphor bronzes is discussed by Kroll, Ball, and Anderson.[35]

Microstructures of a cast phosphorus-treated bronze containing 8% tin are shown in Figs. 69 and 70. The general nature of the cored, dendritic, cast structure is most evident in Fig. 69, in which portions of the alpha dendrites appear in relief. The irregular figures within the dendritic matrix, which are revealed at a higher magnification in Fig. 70, are mixtures of alpha and the brittle compound Cu_3Sn. These mixtures have a eutectoid-like structure * because they were formed, during cooling, from the eutectoid ($\alpha + Cu_{31}Sn_8$). The eutectoid itself (composition 27% Sn) is stable only between 520° C. and 350° C., as shown in Fig. 68.

The presence of the hard, wear resisting Cu_3Sn phase is desirable in cast gear and bearing bronzes, but it is completely absorbed into the alpha solution during processing of the wrought alloys at temperatures within the alpha region of Fig. 68. Although the equilibrium diagram indicates that intermetallic compound particles should precipitate from the alpha solution at low temperatures, the precipitation process is very slow and the second phase is not ordinarily encountered in commercial wrought bronzes. A typical annealed structure of a wrought bronze showing twinned alpha crystals is illustrated in Fig. 71.

The alpha bronzes with low tin contents may be hot reduced but, like the alpha brasses, they are essentially cold working alloys. Most of the production is in the form of sheet and wire for springs,

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* According to Fig. 68, eutectoids form at about 580° C. and 520° C. from beta and gamma solid solutions respectively. Because of the complex nature of this diagram further consideration of eutectoids will be deferred to page 141 (copper-aluminum alloys) and to Figs. 84 and 85 of Chapter VII.

Left. FIG. 69. — PHOSPHOR BRONZE CONTAINING 92% Cu-8% Sn. AS CAST. (Etched.) Rockwell " B " 47. X100.

Right. FIG. 70. — PHOSPHOR BRONZE CONTAINING 92% Cu-8% Sn. AS CAST. (Etched.) Rockwell " B " 47. X500.
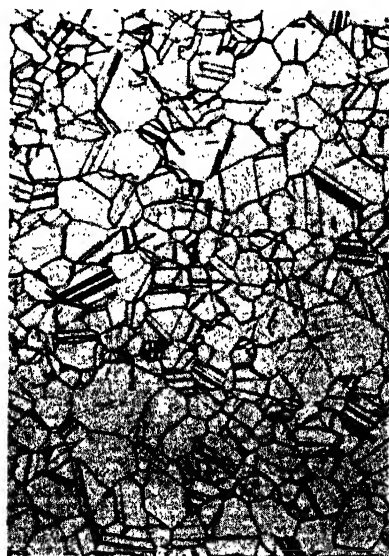
FIG. 71. — WROUGHT PHOSPHOR BRONZE CONTAINING 95% Cu-4.75% Sn-0.25% P. ANNEALED. (Etched.) X75. (*Courtesy of D. K. Crampton.*)

diaphragms, wire cloth, welding rod, electrical and hardware articles, and stamped products including coins. A considerable quantity of 1.5% tin bronze is used as trolley wire having a minimum electrical conductivity equal to 40% that of copper. Bronze tubing is used for handling corrosive fluids and gases.

The following nominal compositions and average tensile strengths are representative of commercial phosphor bronze sheet metals.[36] (More complete data are given in A.S.T.M. Tentative Specifications B103–36T.)

TENSILE STRENGTH — LB./SQ. IN.

% Sn	Soft Temper (0% red. by cold rolling)	Hard Temper (37% red.)	Spring Temper (60% red.)
4–5	47 000	80 000	100 000
8	60 000	92 000	110 000
10	65 000	100 000	122 000

Spring wires are cold drawn to even greater strengths and have good endurance characteristics.

Cast Phosphor Bronzes. — Cast phosphor bronzes containing 10 to 13% tin are known as gun metal, gear bronze, or bearing bronze. The hardness and wear resistance increase with the tin content. High phosphorus content causes the formation of the compound Cu_3P which becomes associated with the Cu_3Sn present and also improves wear resistance.

Phosphor bronze gears wear well against steel and make quiet, efficient assemblies when mated with steel gears. Bronze gear blanks are often cast by the centrifugal process for improved properties.[37]

Bronze Bearing Metals. — The straight phosphor bronze gear and bearing metals withstand much higher unit pressures and higher temperatures than any of the bearing alloys heretofore considered; however, they are commonly modified by additions of zinc and lead to improve their casting properties, plasticity, machinability, and frictional characteristics.

The alloy known as 88–10–2, referring to copper, tin, and zinc respectively, is probably the best known of the lead-free bearing bronzes. It is the original gun metal or government bronze and

still finds wide application because of its excellent combination of strength, toughness, bearing qualities, casting characteristics, and resistance to sea water and other types of corrosion.[38, 39, 40]   It is used for steam valves and fittings, pump parts, bolts, gears, and bushings.

Lead has been found to be almost indispensable for best wearing qualities in bearing bronzes.[41]   Leaded bronzes have complex microstructures containing both plastic and wear resisting phases, and no attempt is made to homogenize their structures by heat treatment.   A well known alloy of this type containing 80% copper, 10% tin, 10% lead (known as 80–10–10) is an excellent bearing metal having greater plasticity, lower coefficient of friction, and better machinability than the lead-free bronzes.[42]   The Brinell hardness of the 80–10–10 alloy is 60 to 70, about twice that of the white bearing metal alloys, but less than the hardness of the lead-free phosphor bronzes.   Despite its lower hardness, the 80–10–10 alloy has been found to outwear the lead-free bronzes.[38]

Leaded bronzes are used extensively for locomotive wearing parts such as rod bushings, cross head jibs, driving boxes, shoes, and wedges.   In addition to the 80–10–10 alloy the following combinations are included in A.S.T.M. Tentative Specifications B66–36T: 7–10% Sn, 10–15% Pb, under 1.25% Zn, balance Cu; 6–8% Sn, 14–22% Pb, under 1.25% Zn, balance Cu; 4–6% Sn, 23–27% Pb, balance Cu.   The plasticity increases as the lead content is increased and the tin is decreased, finally approaching the characteristics of the leaded copper alloys discussed in a previous section of this chapter.

Comparison of White Metal, Copper-Lead, Leaded Bronze, and Silver Engine Bearings. — In commenting on the possibilities of using silver lined bearings for aircraft engines, Zink[43] has summarized the relative performance characteristics of several types of engine bearings as follows:

" Engine tests indicate white metal is a superior bearing material for light loads, low operating temperatures, poor shaft finishes, distorting rubbing surfaces, and lubricants containing a fairly large quantity of abrasives and metallic particles.   By white metal is meant S.A.E. lead or tin base Babbitts or slight modifications

thereof. The ability of white metals to withstand short periods of operation without lubricant has been much overrated. Although white metal bearings do not seize when lubrication is partly interrupted, the result of such interruption shows up later in metal spalling from the backing surface in the region of the poor lubrication. A change in the material for a bearing is often made because of the above condition, the change being made on the false assumption that the material has fatigued. A study of the distribution of the oil over the surface to be lubricated would allow for better performance than could be obtained by the mere substitution of a new, more expensive material that might or might not work under the original conditions of poor lubrication. Whether the new bearing material would work depends to a large extent on how poor the original lubrication was.

" Usually copper-lead has replaced white metal only when higher loads and operating temperatures were encountered. It does not flow as easily as white metal and requires better alignment of journal and bearing surfaces than white metal. The introduction of copper-lead bearings marked the beginning of a period when more attention was given to the hardness of the journal, bearing finishes, and the cleanliness of the oils. Copper-lead was susceptible to corrosion from organic acids in oils after the conditions of operation exceeded a certain temperature. Flaking of the copper-lead from the backing material occurred much the same as in the case of white metal except that, due to the characteristics of copper-lead at temperature, such flaking did not occur until a much higher temperature was reached.

" Copper-lead, if substituted for white metal in the same application and given a proper coating of lubricant, relative freedom from dirt particles in the oil supply, and proper operating clearances from start of operation, will withstand 75 per cent greater loading when the bearing temperature exceeds 150 deg. F.

" Fatigue of both white metal and copper-lead will occur if the lining is caused to flex any appreciable amount. Bronze and silver, because of their greater strength, are relatively free from such troubles.

" Bronze, (high lead bronze such as the compositions (a) 4 per cent tin, 21 per cent lead, copper balance, and (b) 10 per cent tin,

15 per cent lead, copper balance), operates very satisfactorily under extreme high loads and low rubbing velocity, provided journal and bearing alignments are maintained, lubricant is clean, and is distributed properly over the rubbing surfaces. Due to the hardness of these materials they distort only when a tremendous amount of pressure is brought to bear. High pressures when localized, as at the ends of bearings, are sufficiently great to rupture the oil film, producing metal-to-metal contact accompanied by a great amount of heat providing the relative movement of the surfaces is great. Such localized heat distributes itself over the bearing surface, raising the temperature of the surface above the point where the oil will adhere to it. The weaknesses of the bronze as a satisfactory bearing material under conditions of high rubbing velocity, uncertain conditions of distortion and lubrication were at once apparent and a more malleable material was sought. Silver seemed to offer possibilities and a number of engine and bench tests were made of silver bearings.

"Engine tests indicated slightly better performance for silver than bronze although numerous seizures occurred. It was hard to conclude what caused such seizures as no definite changes in fits, methods of lubrication or distortion were apparent. The main unknowns seemed to be surface finishes, and degrees of abrasive or metallic particles in the lubricant. Bench tests, designed to reproduce as closely as possible engine conditions, indicated, first, that silver was more susceptible to seizures than copper-lead or white metal if there was a trace of abrasive or metal particles in the oil; second, that journal finishes were important, lapped journals giving much better results than ground and polished journals; third, grain size of silver was important in reducing seizures."

Self-Lubricating or Oil Impregnated Bearings. — Another comparatively recent development is the self-lubricating or oil-retaining type of bearing alloy. Various mixtures of powdered copper, tin, and graphite are molded under high pressure, sintered at elevated temperatures in a protective atmosphere, and impregnated with oil. As much as 50% of the volume may be air space, capable of taking up oil. The basic composition is often 90% copper-10% tin; however, many other analyses are available.[44]

Powder metallurgy has many other important applications in-

cluding the production of useful products from pure metals and alloys which are too refractory to be melted and cast in the normal way.[45]

Bell Bronzes. — Bell bronzes are cast copper-tin alloys containing 20 to 25% tin. No other composition gives quite the tone quality characteristic of this hard, brittle alloy.

Nickel in Bronzes. — Small amounts of nickel have been found to improve the elastic properties and toughness of many brasses and bronzes, including the gear and bearing bronzes.[46] The effect of nickel upon the constitution of the tin bronzes has been thoroughly investigated.[47]

Silicon Brasses. — The high cost and increasing scarcity of tin is largely responsible for the development of copper alloys containing silicon as a principal alloying element. The most important copper-silicon alloys, the silicon bronzes, are described in the next section. Silicon brasses containing 8 to 15% zinc and 2 to 5% silicon are also available. They have good casting characteristics and mechanical properties, and because of their low cost they can compete with both cast and wrought tin bronzes and with other high strength copper alloys such as manganese bronze.[48] One type containing 14.25% zinc and 4.25% silicon is suitable for die casting.

Silicon Bronzes. — The most important of the copper-silicon alloys are the silicon bronzes containing 1 to 4% silicon with small additions of other elements. The compositions of four established classes of silicon bronzes, as well as the minimum property specifications for rods, are given in Table XIX. In general, the low silicon type, B, is best suited to withstand severe cold deformation in forming or cold heading operations, while the higher silicon alloys develop greater strength upon cold rolling or drawing.

The proprietary alloys such as Everdur, Herculoy, P.M.G., and Duronze generally contain but one or two of the elements listed in addition to copper and silicon; for example, certain Everdur grades contain 1% manganese and P.M.G. contains 1.5% iron.[50] Lead is generally added to the free-cutting alloys for screw machine products.

According to the copper-silicon equilibrium diagram[51] about 4% silicon is soluble in solid copper at room temperature. With
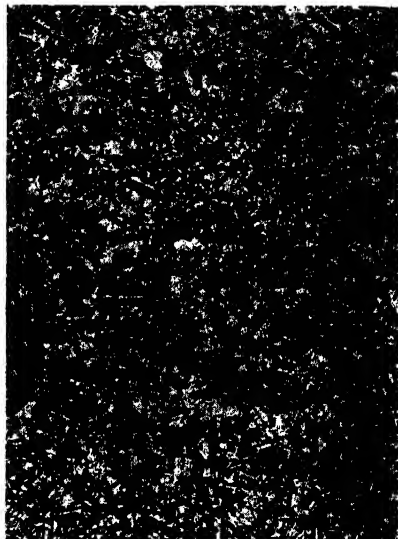
Left. FIG. 72. — WROUGHT SILICON BRONZE. (HERCULOY — 94.8% Cu, 3.2% Si, 1.5% Zn, 0.5% Sn.) COLD DRAWN AND ANNEALED AT 525° C. (997° F.). (Etched.) Brinell 131. X100.

Right. FIG. 73. — WROUGHT SILICON BRONZE. (HERCULOY.) COLD DRAWN AND ANNEALED AT 550° C. (1022° F.). (Etched.) Brinell 124. X100.

FIG. 74. — WROUGHT SILICON BRONZE. (HERCULOY.) COLD DRAWN AND AN-NEALED AT 700° C. (1472° F.). (Etched.) Brinell 88. X100.

TABLE XIX. COMPOSITIONS OF COPPER-SILICON ALLOY RODS, BARS, AND SHAPES

	A	B	C	D
Copper, minimum	94.8	97.0	90.0	90.0
Silicon	2.70–3.75	0.70–1.75	2.30–3.75	1.00–3.75
Lead	—	—	—	0.20–0.60 *

One or more of the following elements may be present within the limits specified below.

	A	B	C	D
Manganese, maximum ...	1.50	0.75	1.50	1.50
Zinc, maximum	1.75	1.25	4.25	4.25
Iron, maximum	1.60	0.80	1.60	1.60
Tin, maximum	0.75	2.25	1.50	2.25
Lead, maximum	0.05	0.05	0.05	—
Sum, minimum	99.5	99.5	99.5	99.5

* or tellurium or selenium.

MINIMUM PROPERTIES OF ½ TO 1 INCH RODS

Temper	Alloy Type	Tensile Strength lb./sq. in.	Yield Point (0.5% set) lb./sq. in.	Elongation in 2 in., %
Soft	A,C,D	52 000	—	45
½ hard	A,C,D	70 000	38 000	25
Hard	A,C,D	85 000	50 000	18
Extra hard	A,C,D	100 000	55 000	10
Soft	B	40 000	—	35
½ hard	B	55 000	20 000	14
Hard	B	65 000	35 000	8
Extra hard	B	75 000	40 000	6

From A.S.T.M. Tentative Specifications B98–36T.[49] Courtesy Am. Soc. for Testing Mats. Bend test requirements are also given in the specification.

the exception of the iron-containing alloys, the silicon bronzes are simple solid solution alloys, hardenable only by cold work. The silicon bronzes have a high capacity for work-hardening — hard rolled or drawn sheet or wire products often attaining strengths well over 100,000 lb. per sq. in.

When heavy cold reductions are made it is sometimes necessary to soften by annealing between cold working operations, as already described in the case of high brasses. A cold drawn rod whose hardness was 228 Brinell was cut into several pieces which were annealed at various temperatures between 100° and 800° C. The

Brinell hardness was 214 after annealing at 350° C. but dropped rapidly at higher annealing temperatures, indicating that recrystallization had occurred. After the 350° C. anneal the microstructure was essentially the same as the very fine textured structure of the cold drawn rod. Furthermore, little change in the microstructure was observed even after annealing at 525° C., as shown by the fine grain structure of Fig. 72. Definite evidence of grain growth is apparent in Fig. 73, obtained after annealing at a temperature only 25° C. higher. Marked grain growth occurred at 600° C. and higher annealing temperatures, as shown in Fig. 74.

These results indicate that changes in mechanical properties are more sensitive than the microstructure in detecting the early stages of recrystallization. This has been found to be true for many other work-hardened alloys.

In addition to their high strength and toughness, the silicon bronzes are characterized by their resistance to many corrosive substances including brines, sulphite solutions, sugar solutions, organic acids, and several non-oxidizing inorganic acids and alkalis.[50] Among the many successful applications of silicon bronzes are processing tanks and equipment, electrical fittings, marine hardware, water wheels, boilers, blowers, pumps, shafting, bolts, etc., operating under corrosive conditions.

Many of the above mentioned products are made from castings or combinations of cast and wrought parts. Silicon bronzes are not difficult to cast compared with other high strength alloys such as aluminum bronzes. One of the better known casting alloys contains 95% copper, 4% silicon, and 1% manganese; however, higher silicon contents are sometimes used. Bedworth and Weaver [52] give the following approximate mechanical properties for sand castings of this alloy:

Tensile strength, lb./sq. in.	50 000
Yield strength at 0.75% set, lb./sq. in.	20 000
Elongation in 2 in., %	20
Reduction in area, %	25
Brinell hardness (500 kg.)	80

Copper-Nickel Alloys. — The simplicity of the copper-nickel equilibrium diagram, compared with other copper alloy systems, is

evident in Fig. 75. All of the alloys are face-centered cubic solid solutions and they exhibit the same dendritic coring, work-hardening, recrystallization, and twinning effects observed in other alpha solid solution copper alloys.[53]

A familiar alloy, the United States nickel coin, is stamped from sheets containing 75% copper-25% nickel. Its white color is characteristic of all the copper alloys containing 20% nickel or over.

FIG. 75. — THE COPPER-NICKEL EQUILIBRIUM DIAGRAM. (*Metals Handbook*, 1939, *p. 1353*.)

The 30% nickel alloy has proved to be the best available for the severe service conditions to which marine steam condensers are subjected. Although more expensive, it is preferred over Admiralty metal and related alloys for resisting the corrosive action of salt water passing through the hot tubes at very high velocities. Worthington[54] describes the mechanism of corrosion under these conditions, as well as the alloys used. The 30% nickel alloy has been applied to other equipment requiring high strength at moderately elevated temperatures, as well as resistance to hot and cold brines, sodium hydroxide, chlorine bleaching solutions, and other corrosive substances. A modification containing 29% nickel and 1% tin is used for similar applications.[55]

In addition to tubes and sheets, castings are required for fittings, valves, pump bodies, etc., in most installations. Pressure tight castings of a modified analysis are available.[56]

Monel Metal. — Monel metal is an alloy smelted from a mixed nickel-copper ore mined in Sudbury, Ontario. It contains nominally 67% Ni, 30% Cu, balance Fe, Mn, Si, and C; however, its composition is modified for specific applications. In the wrought form it is a face-centered cubic alpha solid solution alloy having the usual type of homogeneous grain structure already observed in Figs. 51 to

58, 71, 73, and 74. Monel metal is produced as sheet, rod, and wire, as well as in cast form. It is used for a wide variety of applications requiring high strength, pleasing appearance, good elevated temperature properties, and resistance to many corrosive substances.

Annealed Monel metal rod has a tensile strength of about 80,000 lb. per sq. in. with 45% elongation in 2 inches and very high impact toughness. The corresponding properties for cold drawn rod are approximately 100,000 lb. per sq. in. and 25% elongation.[57]

Monel metal retains its strength at elevated temperatures to a greater extent than the copper-rich alloys, which enables it to compete with heat and corrosion resisting steels for certain types of tanks, boilers, piping, etc., for processing equipment.

Among the applications of Monel metal which depend mainly on its corrosion resistance are: sinks and other kitchen equipment; containers, pumps, valves, etc., for the food, textile, petroleum, and paper making industries; and equipment for handling steel in pickling operations.

A modified wrought alloy known as K-Monel contains 3 to 5% aluminum. It is hardenable to 350 Brinell by a precipitation heat treatment, but is more often used at about 275 Brinell and 140–150,000 lb. per sq. in. tensile strength with 30% elongation in 2 inches.[57, 58] K-Monel has approximately the same corrosion resistance as ordinary Monel, and has the advantage of developing its high mechanical properties by heat treatment instead of by cold working. This is often an important consideration in that all fabrication may be completed before hardening the alloy.

A free-machining grade known as R-Monel is also available.[57]

Commercially pure nickel (about 99.5% Ni), known as malleable nickel or A-Nickel [57], also has excellent mechanical and chemical properties, and although it is more expensive than Monel, it is used for similar purposes where special corrosion resisting qualities are required.

Both nickel and Monel are produced in cast form for pipe fittings, propellers, pump casings, valve bodies, agitators, etc.[59] The following properties for Monel metal sand castings are given by Murphy [28]:

Tensile strength, lb./sq. in. 53 700–76 200
Yield point, lb./sq. in 21 360–40 300
Elongation in 2 in., % 30–15
Brinell hardness 120–160

The addition of silicon up to about 4% is said to improve the casting properties of Monel metal. It also increases hardness and strength at the expense of ductility. The silicon-Monel castings are amenable to precipitation hardening by a quenching and tempering procedure.[28]

Nickel and the Monel metal compositions are discussed in detail in various " Metals Handbook " articles.

Other Nickel Alloys. — Nickel is the base for a number of important alloys which do not contain copper. Although a chapter might be devoted to the nickel alloys, it will only be possible to call attention to a few well known compositions in this text.

In the majority of cases the high nickel alloys are characterized by heat and corrosion resistance. For example, Inconel, containing approximately 80% nickel with 15% chromium and 5% iron in solid solution, has been found useful for aircraft engine exhaust equipment because of its good mechanical properties at elevated temperatures and its resistance to hot combustion gases as well as to atmospheric corrosion.

A somewhat similar alloy with 60% nickel, 15% chromium, and 25% iron is widely used in the form of wire for electrical resistance elements operating at temperatures up to 900° C. (1652° F.) and for other elevated temperature applications. An iron-free alloy with 80% nickel and 20% chromium is used for similar purposes at temperatures up to 1100° C. (2012° F.).

Konel [60] is a precipitation hardening alloy containing 73% nickel, 17% cobalt, 7.5% iron, and 2.5% titanium. In addition to its good stability at elevated temperatures, Konel develops high strength and hardness which it retains up to at least 600° C. (1112° F.). Among its applications are the heated elements of vacuum tubes.

Nickel-Silvers. — Nickel-silvers, also called German silvers, are sometimes classified as brasses in which some of the zinc has been replaced by nickel. The compositions vary widely within the limits

10 to 30% nickel, 5 to 50% zinc, balance copper, a typical composition being 65% copper, 17% zinc, 18% nickel.[61]   All but the very high zinc types are solid solution alloys having face-centered cubic crystal structures.   They are homogeneous in the wrought condition with microstructures similar to those in Figs. 51 to 58.

Nickel-silvers have a white color similar to silver and are highly resistant to corrosion by foodstuffs and the atmosphere. They are used extensively for silver plated tableware, hollowware, jewelry, and related products.   In wrought form they are also used for springs, electrical resistance wire, key stock, architectural trim, and food handling equipment.   Nickel-silver castings are used for domestic plumbing fixtures and marine fittings.

Aluminum Bronzes. — The aluminum bronzes are copper-aluminum alloys with additions of iron, manganese, or nickel.

FIG. 76. — THE COPPER-ALUMINUM EQUILIBRIUM DIAGRAM. (Modified form of Stockdale's diagram.)

In this country iron is widely used as the third element because of its beneficial effect on the mechanical properties.

The copper-aluminum equilibrium diagram, Fig. 76, indicates solubility of aluminum in solid copper up to 9.8%.   Most of the wrought aluminum bronzes are homogeneous alpha solid solution alloys with compositions within this range.   The cast aluminum bronzes have aluminum contents up to 13.5%, and their microstructures contain a eutectoid constituent which is similar to the eutectoid known as pearlite in carbon steels.   (See page 159.)

The copper-rich eutectoid alloy of the copper-aluminum system contains 11.9% aluminum.   It consists of a mechanical mixture of the alpha and delta $(\alpha + \delta)$ solid solutions.   In many respects

eutectoid mixtures are similar to eutectics, both in microstructure and the manner in which they are formed. Eutectic mixtures, however, form from liquid solutions (see Fig. 20, for example) whereas eutectoids form upon cooling solid solutions. The $(\alpha + \delta)$ eutectoid is the result of a constant temperature transformation of the beta solid solution at 568° C. As in the case of eutectic systems, alloys which depart somewhat from the true eutectoid composition precipitate a primary phase before the eutectoid transformation temperature is reached. The primary alpha solid solution which would form between 700° C. and 568° C. upon cooling an 11% aluminum alloy is, therefore, comparable to the primary bismuth of Fig. 21. The eutectoid constituent itself is generally a finely divided mixture of the two phases present. (See Fig. 77 and Figs. 88 to 95 of Chapter VII.)

The aluminum bronzes have many outstanding characteristics including high strength and hardness in the cast state, capacity for hardening and strengthening by heat treatment, exceptionally good corrosion resistance, good retention of hardness and scale resistance at elevated temperatures, good wearing qualities when used with steel, and lower density than other copper alloys.

Although the cost of the raw materials is less than for tin bronzes, the foundry costs are usually high because of special melting and casting requirements.[62, 63] The low density of the aluminum or aluminum-iron additions to the charge make it difficult to obtain homogeneous castings without pigging and remelting. Oxide films form readily upon exposure to air, consequently turbulent pouring must be avoided. Aluminum bronzes also undergo high shrinkage upon solidification and upon further cooling, therefore adequate risers and suitable patterns must be provided. The use of chills in the molds to control the solidification process is also helpful. The high shrinkage of these bronzes does not ordinarily cause cracking because they have good strength at elevated temperatures; in fact, the 10% aluminum-1% iron alloy is successfully die cast.

Ingots for rolling to bars, plate, sheet, or strip generally contain 4 to 7% aluminum. The strength of the wrought alloys increases with aluminum content, ranging from 50–80,000 lb. per sq. in. with 40 to 15% elongation in 2 inches in the forged or rolled condition. Much higher strengths are obtained by cold rolling.[64] The micro-

structures of these alloys are similar to those of other wrought, copper base, alpha solid solution alloys. The applications of these bronzes depend largely upon their resistance to corrosive atmospheres, sea water, steam, combustion products, and many neutral and acid salt solutions.

The casting alloys have higher aluminum contents which increase their hardness and wear resistance. Typical compositions and applications are given in Table XX.

In general, the tensile strengths of these castings increase along with the hardness, ranging from about 55–95,000 lb. per sq. in. with 25 to 1% elongation in 2 inches. The 11.5% aluminum alloy, for example, has the following properties in the cast state: 75–85,000 lb. per sq. in. tensile strength; 33–42,000 lb. per sq. in. yield point; 14 to 10% elongation in 2 inches; and 167 to 179 Brinell.

FIG. 77. — CAST ALUMINUM BRONZE CONTAINING 86.9% Cu, 10.7% Al, 2.4% Fe. (Etched.) X100.

The microstructure of a cast aluminum bronze is shown in Fig. 77. The white constituent is alpha solid solution and the dark constituent is the eutectoid $(\alpha + \delta)$. The small dark particles are an intermetallic compound generally believed to be $FeAl_3$. The compound particles and the eutectoid, which are hard and wear resistant, increase in amount as the aluminum and iron contents are increased. As indicated in the footnote to Table XX, the microconstituents obtained in commercial castings depend upon the foundry practice. It is also apparent that these structures do not conform strictly to the copper-aluminum equilibrium diagram since the alpha constituent is present in the 12.5% aluminum alloy, which should consist of delta plus eutectoid according to the diagram.[65]

It is possible to improve the hardness and other properties of

TABLE XX. TYPICAL COMPOSITIONS AND APPLICATIONS OF CAST ALUMINUM BRONZES

Per Cent Composition			Micro-structure	Brinell As Cast	Typical Applications
Cu	Al	Fe			
89.0	10.0	1.0	varies with treatment	100	Heat treated castings for machine parts, marine service, ordnance, acid equipment, and general purposes.
88.0	9.0	3.0	alpha	120	Propellers and other marine castings, pump parts, bushings and bearings for soft steel shafts, gears for light service. (Also rolled or forged valve stems, bolts, pump rods.)
86.0	10.5	3.5	alpha + eutectoid (mostly alpha)*	140	Bushings and bearings for machines and machine tools.
84.5	11.5	4.0	alpha + eutectoid *	170	Gears and heavy duty bearings, acid resisting equipment, welding jaws, non-sparking tools.
83.0	12.5	4.0	alpha + eutectoid (mostly eutectoid)*	235	Bushings and bearings for hardened steel shafts. Forming dies for light gage mild steel.
82.5	13.0	4.0	delta + eutectoid *	295	Drawing and forming dies for steel. Bearing plates, liners, guides, etc., for use against hardened steel.
82.0	13.5	4.5	delta + eutectoid *	325	Drawing and forming dies for heavy gage and stainless steels. Chisels and other non-sparking tools requiring high hardness.

* These microconstituents are present in cast Ampco Metal. Variations are possible depending upon the melting and casting practice and the use of special elements in the composition. In all cases an intermetallic compound containing iron is also present.

many of the aluminum bronzes by transformation to the beta phase at elevated temperatures, followed by rapid cooling and reheating to temperatures below the eutectoid transformation. The normal crystal structure of the beta phase, as it exists at elevated temperatures, is modified somewhat upon cooling rapidly. Reheating the quenched alloy causes separation of fine alpha and delta particles. Depending on the reheating temperature various combinations of properties are obtainable. The 10% aluminum-1% iron alloy is ordinarily used for heat treated castings.

The hardness and wear resistance of certain aluminum bronzes make them particularly suitable for operation with steel as gears and bearings. The hardness of the delta plus eutectoid types is sufficiently high to withstand deformation in service as forming dies for heavy gage steels and for stainless steels. In these applications the problem of lubrication to prevent seizing, "pick-up," and scratching is much less difficult than it is when using hardened tool steel dies. This is particularly important in forming and drawing stainless steel. Automobile frame members, fenders, headlamps, washing machine tubs, aluminum window frames, stainless steel dairy equipment, nickel-silver hollowware, and many other products are successfully formed or drawn in aluminum bronze dies.[66] The brittleness of these alloys and their inability to maintain a cutting edge prevents their application for blanking or shearing dies; nor are they recommended for hot forging operations or for working copper, brass, or bronze. The hardened tool steel dies (see Chapter XV) have a wider field of application and are, of course, used in much greater quantities than bronze dies.

Bronze hand tools reduce the danger of fires and explosions caused by sparks, consequently they are widely used in the oil fields and refineries and in other industries handling inflammable or explosive gases and liquids. Certain aluminum bronzes, as noted in Table XX, are sufficiently hard and strong for this type of service.

Although aluminum bronzes cannot compete with steel for forging dies, they have better heat resistance and freedom from warping at elevated temperatures than other bronzes. Heat treating temperatures of 1700° F. do not scale the surface, probably because of the formation of a very thin protective film of Al_2O_3. Valve seat

inserts and other products requiring heat resistance have been made of these alloys.

The gold color and resistance to staining of certain aluminum bronzes has resulted in special applications such as Dirigold tableware.

Precipitation Hardening Copper Alloys. — In addition to the many hardened copper alloys already discussed in this chapter, there are several other alloys which depend upon precipitation heat treatments to develop their best properties. Included in this classification are beryllium bronzes, the copper-nickel-silicon bronzes which may be hardened to over 200 Brinell by precipitation of the compound Ni_2Si [67], and a copper-chromium-silver alloy (containing over 99.5% copper) recently developed for welding electrodes and other electrical applications.[68]

Copper-Beryllium Alloys. — The art of hardening copper reached a new high level with the discovery of the precipitation hardening characteristics of copper-beryllium alloys by Masing in 1926. These alloys may be hardened to over 400 Brinell; however, their most important commercial applications depend on the excellent combinations of strength, ductility, and electrical conductivity developed at somewhat lower hardnesses.

The element beryllium has approximately the same density as magnesium; however, it is of little interest as a base for light alloys because of its brittleness and prohibitive cost. In the form of a master alloy for making beryllium bronze, the cost is $23 per pound of contained beryllium with the copper at market price (Jan. 1939). The great difficulty of reducing the ores [69] makes it doubtful whether the cost of the beryllium will ever approach that of tin, nickel, or aluminum, even though the output is greatly increased. Fortunately, small amounts of beryllium are sufficient to harden the commercial alloys, the amounts used ranging from about 2.25% beryllium in binary copper alloys to 0.1% in certain ternary copper base alloys.

Borcher's equilibrium diagram, reproduced in Fig. 78, clearly indicates the possibility of precipitation hardening, based on the decreasing solubility of beryllium in copper with decreasing temperature. The usual quenching temperature for the solution heat

treatment is 800° C. The quenched alloys retain the beryllium in the alpha solid solution phase. (High beryllium alloys will be mixtures of alpha and beta.) Quenching produces maximum softness, and annealing by cooling slowly from high temperatures is not recommended. Reheating the quenched alloys to temperatures be-
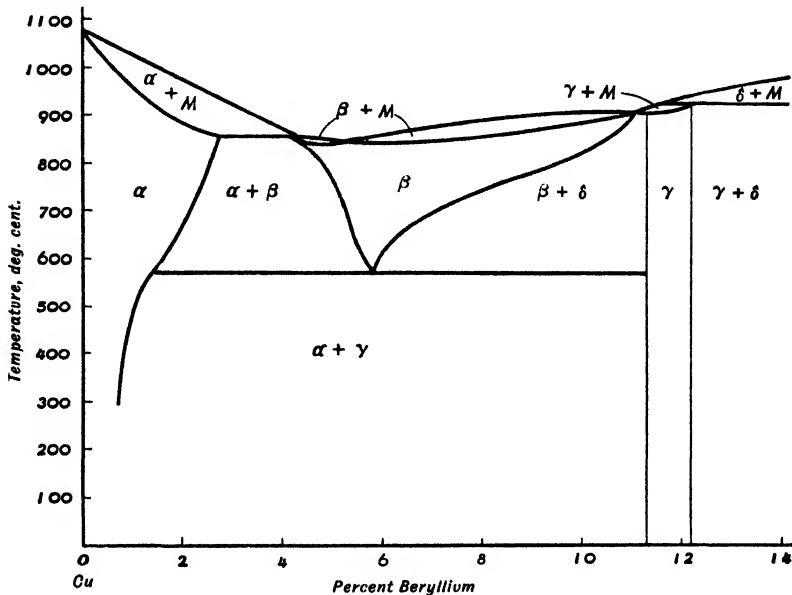
FIG. 78. — THE COPPER-BERYLLIUM EQUILIBRIUM DIAGRAM. (Borcher's Diagram. Courtesy of E. S. Rowland and C. Upthegrove, Trans. Am. Inst. Mining and Met. Eng.[70])

tween 250° and 350° C. precipitates the hard delta phase from the supersaturated alpha solid solution.

When fabricating operations permit, as in wire drawing, the quenched alloys may be work-hardened either before or after the reheating treatment to obtain the highest possible hardness.[71]

The compositions and properties of typical beryllium bronzes in the fully heat treated condition are given in Table XXI.

The binary alloys with 2.0 to 2.75% beryllium are used for springs, contact clips, breaker points, diaphragms, commutator segments, and other parts of electrical machinery requiring high mechanical properties, including fatigue and wear resistance, and mod-

erately high electrical conductivity. These alloys begin to soften when used above their hardening temperature range of 250° to 350° C. Alloys 4, 5, and 6 of Table XXI, which were developed specially for welding electrodes, harden by precipitation at 450° to 500° C., therefore they retain their properties at higher tempera-

TABLE XXI. COMPOSITIONS AND PROPERTIES OF TYPICAL
PRECIPITATION HARDENED BERYLLIUM BRONZES

	1	2	3	4	5	6
Copper, %	97.81	97.5	97.25	97.0	99.5	99.5
Beryllium, %	2.19	2.5	2.75	0.4	0.1	0.1
Third element, % .	—	—	—	2.6 Co	0.4 Cr	0.4 Cr
Cast or wrought ..	cast	wrought	cast	cast	cast	*
Proportional limit, lb./sq. in.	—	—	—	45 000	15 500	27 000
Yield strength, lb./sq. in.	86 200	88 700	135 000	—	—	—
Tensile strength, lb./sq. in.	110 000	123 500	150 000	90 000	32 500	46 200
Elongation in 2 in., %	1	11	1	10	12	20
Brinell hardness ..	375	365	390	220	80	108
Elec. conductivity, % I.A.C.S.	—	32	—	47	73	74
Reference	69	69	72	71	71	71

* Cast as $1\frac{1}{4}$ inch diameter rods, heated to 900° C. for 1 hour, quenched in water, cold swaged to 1 inch diameter, reheated $1\frac{1}{2}$ hours at 500° C.

tures than the binary beryllium bronzes. Other ternary alloys containing zirconium and titanium are available for elevated temperature applications.

When strength and wear resistance rather than high electrical conductivity is required, the hard binary alloys may be used as forming and drawing dies, molds for plastics, non-sparking tools, springs, worm gears, and bearing metal. The mechanical properties available surpass even those of the aluminum bronzes for certain purposes.

Beryllium is used as an alloying element in nickel, iron, and aluminum alloys; however, only the nickel alloys approach the beryllium bronzes in commercial importance.

Although this treatment of nonferrous alloys is necessarily incomplete, the basic principles of working and heat treating these alloys have been considered. Stellite and cemented carbide tools are important nonferrous alloys to be discussed later as cutting tools. Certain heat resisting alloys of a nonferrous nature are mentioned in Chapter XIV.

REFERENCES

1. Woldman, N. E., and Dornblatt, A. J., " Engineering Alloys," Am. Soc. for Metals, Cleveland (1936)
2. Smith, C. S., and Palmer, E. W., " Thermal and Electrical Conductivities of Copper Alloys," Trans. Am. Inst. Mining and Met. Eng., **117**, 225 (1935)
3. Rhines, F. N., and Mathewson, C. H., " Solubility of Oxygen in Solid Copper," Trans. Am. Inst. Mining and Met. Eng., **111**, 337 (1934)
4. Zickrick, L., " Properties of Copper Deoxidized with Calcium," Trans. Am. Inst. Mining and Met. Eng., **104**, 152 (1933)
5. Cone, E. F., " Oxygen-Free High-Conductivity Copper," Metals and Alloys, **8**, 33 (1937)
6. Fuller, T. S., " The Embrittlement of Solid Copper Containing Oxygen in Hot Reducing Atmospheres," Metal Progress, **31**, 627 (1937)
7. Hansen, E. K., and Brace, P. H., " Furnaces and Atmospheres for Annealing Copper with Electric Heat," Metals and Alloys, **8**, 215 (1937)
8. Webber, H. M., " Copper Brazing in Controlled Atmosphere Furnaces," Metals and Alloys, **2**, 284 (1931)
9. Jenkins, F. G., " Copper-lead Bearing Metals Favored for Severe Service," Metal Progress, **32**, 145 (Aug. 1937) — (See also Sept., Nov., and Dec. 1937)
10. Osborg, H., " Minor Constituents in Copper-Lead Alloys," Metal Progress, **33**, 43 (1938)
11. Fetz, E., " Copper-Lead Bearings from Metal Powder," Metals and Alloys, **8**, 257 (1937)
12. Saeger, C. M. Jr., " Studies on Cast Red Brass for the Establishment of a Basic Classification of Non-Ferrous Ingot Metals for Specification Purposes," Trans. Am. Foundrymen's Assoc., **42**, 67 (1934)
13. Price, W. B., " Properties of Wrought Copper-Zinc-Tin Alloys," Metals Handbook, 1463, Am. Soc. for Metals, Cleveland (1939)
14. Bassett, W. H., and Davis, C. H., " Physical Characteristics of Commercial Copper Zinc Alloys," Trans. Am. Inst. Mining and Met. Eng., **78**, 55 (1928)
15. Pratt, R. S., " Hot and Cold Working of Copper Alloys," Metals Handbook, 1480, Am. Soc. for Metals, Cleveland (1939) — and — " Annealing of Copper Alloys," Metals and Alloys, **6**, 1 (1935)
16. Pratt, R. S., " Properties of Wrought Copper-Zinc Alloys," Metals Handbook, 1399, Am. Soc. for Metals, Cleveland (1939)
17. Mathewson, C. H., " Twinning in Metals," Trans. Am. Inst. Mining and Met. Eng., **78**, 7 (1928)
18. Jeffries, Z., and Archer, R. S., " The Science of Metals," McGraw-Hill Book Company, New York (1924)

19. Crampton, D. K., " Brass for Severe Requirements," Metal Progress, **29,** 39 (May 1936)

20. Crampton, D. K., " Internal Stress and Season Cracking in Brass Tubes," Trans. Am. Inst. Mining and Met. Eng., **89,** 233 (1930) — and — Morris, A., " Stress-corrosion Cracking of Annealed Brasses," 256

21. Jennison, H. C., " Mercurous Nitrate Test," Metals and Alloys, **6,** 348 (1935)

22. Mitchell, N. W., " Corrosion Resistance of Copper Alloys," Metal Progress, **28,** 38 (Dec. 1935)

23. Morris, A., " Machinability of Free-cutting Brass Rod," Trans. Am. Inst. Mining and Met. Eng., **99,** 323 (1932)

24. Crampton, D. K., and Croft, H. P., " Free-Cutting Brass," Metal Progress, **23,** 31, (Jan. 1933)

25. Williams, R. S., and Homerberg, V. O., " Principles of Metallography," 80, McGraw-Hill Book Company, New York (1935)

26. Freeman, J. R., " Die-casting of Brass," Trans. Am. Inst. Mining and Met. Eng., **117,** 371 (1935)

27. Ellis, O. W., " High-strength Brasses," Trans. Am. Inst. Mining and Met. Eng., **83,** 508 (1929)

28. Murphy, A. J., " Non-Ferrous Casting Alloys of High Strength," Trans Am. Foundrymen's Assoc., **43,** 369 (1935)

29. Hensel, F. R., " Manganese Bronze — A Correlated Abstract," Trans. Am. Foundrymen's Assoc., **44, 457** (1936)

30. Eyles, A., " Bronze Propellers for the Greatest Liner," Metal Industry, New York, **32,** 385 (1934)

31. Duma, J. A., " The Manufacture of Cast Trunnion Bearing Metal — Part I," Metals and Alloys, **9,** 111 and 139 (1938)

32. Heycock, C. T., and Neville, F. H., " On the Constitution of the Copper-Tin Series of Alloys," Philosophical Transactions, **202,** Series A, 1 (1904)

33. Greninger, A. B., and Mooradian, V. G., " Strain Transformation in Metastable Beta Copper-zinc and Beta Copper-tin Alloys," Trans. Am. Inst. Mining and Met. Eng., **128,** 337 (1938)

34. Dews, H. C., " The Metallurgy of Bronze," Sir Isaac Pitman and Sons, London (1930)

35. Kroll, L., Ball, F. E., and Anderson, E. A., " Melting of Wrought Phosphor Bronze," Metals and Alloys, **8,** 307 (1937)

36. " Tentative Specifications for Sheet and Strip Phosphor Bronze, A. S. T. M. Designation B103–36T," Proc. Am. Soc. for Testing Mats., **36,** Part I, 734 (1936)

37. Rowe, F. W., " Centrifugal Bronze Gear Blanks," Metal Progress, **23, 15** (June 1933)

38. Harder, O. E., and Cole, C. S., " A Study of Six Bearing Bronzes," Trans. Am. Foundrymen's Assoc., **4,** 314 (1933)

39. Fawcett, L. H., " Properties of Copper-Tin-Zinc Alloys," Metals Handbook, 1452, Am. Soc. for Metals, Cleveland (1939)

40. Jenkins, F. G., " Zinc Bronze — Composition G," Metals and Alloys, **7,** 279 (1936)

41. Staples, E. M., Dowdell, R. L., and Eggenschwiler, C. E., " Bearing Bronzes with Additions of Zinc, Phosphorus, Nickel, and Antimony," Bur. of Stds. Jnl. Research, **5, 349** (1930)

42. Clamer, G. H., "Properties of Cast Copper-Lead-Tin Alloy," Metals Handbook, 1428, Am. Soc. for Metals, Cleveland (1939)

43. Dayton, R. W., "The Possibilities of Silver-Rich Alloys as Bearing Materials," Metals and Alloys, **9** — discussion by George A. Zink on page 328 (1938)

44. Chase, H., "Minimizing Bearing Troubles with Self-Lubricating Bronze Bushings," Machinery, New York, **41**, 9 (Sept. 1934)

45. Noel, D. O., Shaw, J. D., Gebert, E. B., "Production and Some Testing Methods of Metal Powders," — and — Comstock, G. J., "Types of Metal Powder Products — a Classification," Trans. Am. Inst. Mining and Met. Eng., **128**, 37 (1938)

46. Pilling, N. B., and Kihlgren, T. E., "Some Effects of Nickel on Bronze Foundry Mixtures," Trans. Am. Foundrymen's Assoc., **39**, 93 (1931)

47. Eash, J. T., and Upthegrove, C., "The Copper-rich Alloys of the Copper-nickel-tin System," Trans. Am. Inst. Mining and Met. Eng., **104**, 221 (1933)

48. Richards, E. F., "Silicon Brasses: A New Series of Technical Alloys," Metal Industry, London, **44**, 269 (1934)

49. "Tentative Specifications for Copper-Silicon Alloy Rods, Bars, and Shapes — A.S.T.M. Designation B98-36T," Proc. Am. Soc. for Testing Mats., **36**, Part I, 747 (1936)

50. Thum, E. E., "New Copper-Silicon Alloys; History and Foundry Practice," Metal Progress, **33**, 258 (1938) — and — **33**, 389 (1938)

51. Smith, C. S., "Constitution of Copper-Silicon Alloys," Metals Handbook, 1356, Am. Soc. for Metals, Cleveland (1939)

52. Bedworth, H. A., and Weaver, V. P., "Sand Castings of Copper-Silicon Alloys," Trans. Am. Foundrymen's Assoc., **44**, 193 (1936)

53. Jennison, H. C., and Girvin, W. S., "Properties of Copper-Nickel Alloys," Metals Handbook, 1415, Am. Soc. for Metals, Cleveland (1939)

54. Worthington, R., "Copper-Nickel Tubes, Their Advantages for Steam Condensers," Metal Progress, **19**, 20 (July 1933)

55. Price, W. B., "Adnic, A White Metal Alloy for Corrosion-resistance and for Moderate Temperature Heat-resistance," Metals and Alloys, **5**, 77 (1934)

56. Kihlgren, T. E., "Production of Pressure Tight Castings in 30 per cent Cupro-Nickel," Trans. Am. Foundrymen's Assoc., **45**, 225 (1937)

57. Catlin, C. F., and Mudge, W. A., "Compression, Impact, and Other Mechanical Properties of Nickel and Some of Its High Strength, Corrosion-Resisting Alloys," Proc. Am. Soc. for Testing Mats., **38**, Part II, 269 (1938)

58. "Engineering Properties of 'K Monel' Metal," Metals and Alloys, **6**, 101 (1935)

59. Wheeler, E. S., "Monel Metal and Nickel Foundry Practice," Trans. Am. Inst. Mining and Met. Eng., **89**, 458 (1930)

60. Austin, C. R., "High Temperature Properties of Nickel-Cobalt-Iron Base Age-Hardening Alloys — Part I," Trans. Am. Soc. for Metals, **24**, 451 (1936)

61. Kihlgren, T. E., and Wise, E. M., "Physical and Casting Properties of the Nickel Silvers," Trans. Am. Inst. Mining and Met. Eng., **117**, 279 (1935)

62. Crown, J. E., "Founding of Aluminum Bronze," Trans. Am. Foundrymen's Assoc., **44**, 359 (1936)

63. Ganzauge, M. T., "Production of Aluminum Bronze Castings to Withstand High Pressure," Trans. Am. Foundrymen's Assoc., **44**, 479 (1936)

64. Strauss, J., and Fawcett, L. H., "Properties of Wrought Copper-Aluminum Alloys," Metals Handbook, 1408, Am. Soc. for Metals, Cleveland (1939)

65. Bray, J. L., Carruthers, M. E., and Heyer, R. H., " The Coefficient of Equivalence of Iron with Respect to Aluminum in Aluminum Bronze," Trans. Am. Inst. Mining and Met. Eng., **122,** 337 (1936)

66. Zaiser, J. D., and Dreher, G. K., " Bronze, Its Use as a Forming and Drawing Die Material," Iron Age, **136,** 22 (July 25, 1935)

67. Gonser, B. W., and Van Wert, L. R., " The Age-Hardening Characteristics of Some Copper-Nickel-Silicon Alloys," Metals and Alloys, **5,** 251 and 281 (1934)

68. Brace, P. H., " Composition and Properties of Cupaloy in the ' Time Capsule,' " Metals and Alloys, **9,** 311 (1938)

69. Delmonte, J., " Beryllium and its Alloys," Metals and Alloys, **7,** 175, 211, 239 (1936)

70. Rowland, E. S., and Upthegrove, C., " Equilibrium Relations in the Copper Corner of the Ternary System Copper-tin-beryllium," Trans. Am. Inst. Mining and Met. Eng., **117,** 190 (1935)

71. Harrington, R. H., " Precipitation-hardening and Double Aging," Trans. Am. Inst. Mining and Met. Eng., **124,** 172 (1937)

72. Stott, L. L., " Properties and Alloys of Beryllium," Trans. Am. Inst. Mining and Met. Eng., **122,** 57 (1936)

CHAPTER VII

IRON AND CARBON STEELS

Low carbon steels are produced in much larger quantities than all nonferrous metal products combined. Other ferrous metals to be studied in this and the remaining chapters include commercial irons, medium carbon, high carbon, and tool steels, alloy steels, cast iron, and cast steel.

Iron. — Test results for commercially pure irons were used in Chapter I to demonstrate the principles of strain-hardening and recrystallization. In many other respects the properties of iron are comparable to those of other common plastic metals.

The microstructure of iron recrystallized at 1200° F. (Fig. 14) consists of a simple grain structure. Iron is similar to aluminum but dissimilar to copper and nickel in that it seldom forms annealing twins; however, mechanical twins or Neumann bands are not unusual. Neumann bands are regions within a crystal in which an orientation change has taken place caused by mechanical deformation. (See page 6.) In the case of pure iron and very low carbon steels a sudden deformation, as under a hammer blow, is necessary to produce mechanical twinning.

FIG. 79. — NEUMANN BANDS IN PURE IRON. (Etched with 3% nital.) X250. (*Courtesy of J. E. Anderson.*)

The bands in Fig. 79 were found directly below a punched stencil mark in a bar of iron. Neumann bands are of little significance in iron except as an indication of shock loading.

Commercial Irons. — Open hearth pure iron or "ingot iron" developed by The American Rolling Mill Co., is a commercially pure iron available as sheet, strip, plate, rods, wire, and castings. Galvanized ingot iron sheets are used for drainage culverts and flumes, roofing and siding, roof drainage parts, furnaces, ducts, farm equipment, and numerous other applications. Other materials used for the same purposes are galvanized low carbon and low alloy steels, and in some cases sheet copper. Ingot iron sheets are also used very extensively for porcelain enamel applications such as refrigerator cabinets, stoves, washing machine tubs, sign panels, etc., because of their excellent enameling characteristics as well as deep drawing properties. (See also Chapter X.) Typical applications of uncoated or black sheets and plates are pipe, tanks, stacks and breeching, grave vaults, barges and other boats, and plates for low pressure boilers.

Fig. 80. — Ingot Iron Enameling Sheet. Annealed. (Etched with 3% nital.) Rockwell "B" 38. X100. (*Courtesy of D. J. Fergus.*)

A typical Armco ingot iron contains 0.015% carbon, 0.025% manganese, 0.003% silicon, 0.005% phosphorus, 0.025% sulphur. The tensile properties in the annealed condition are approximately: 45,000 lb. per sq. in. tensile strength; 30,000 lb. per sq. in. yield strength; 35 to 40% elongation in 2 inches; and 70 to 75% reduction in area. The microstructure of an enameling iron is shown in Fig. 80. The rounded inclusions in the ferrite grains are mainly iron oxide.*

Commercially pure irons have certain electrical applications.

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* Nital (a dilute solution of nitric acid in alcohol) is generally used for etching iron and steel structures; however, picric acid in alcohol, which acts much slower, is often preferred to show detail in the constituent known as pearlite.

Forged or wrought forms are used for the iron portions of direct current magnetic circuits in which high permeability and low remanence are required. Annealing at high temperatures improves these properties. A typical microstructure of an electrical grade of ingot iron, annealed at 1700° F., is shown in Fig. 81. In alternating current applications the high conductivity of relatively pure iron results in excessive eddy current losses, hence iron-silicon alloys are generally used.

Irons similar in composition to Armco ingot iron, but specially treated to produce very dense castings, have been used for large magnet parts where high permeability at high induction is important.

The magnetic properties of commercial and high purity irons are included in the comprehensive treatise, " The Metal Iron." [1]

Wrought Iron. — Before the development of the Bessemer process for making steels on a high tonnage basis, hand processed wrought iron was the principal ferrous metal used for constructional and machine purposes. It is still manu-

FIG. 81. — INGOT IRON SHEET FOR ELECTRICAL PURPOSES. ANNEALED AT 1700° F. (Etched with 3% nital.) Rockwell " B " 22. X100. (*Courtesy of W. L. Havekotte.*)

factured, both by hand puddling and by a comparatively new mechanical process, and is used mainly for welded pipes and for structural plates, shapes, and bars for farm implements, architectural, railroad, and shipbuilding purposes. In making wrought iron considerable quantities of slag, usually 2% and up to 4%, are mechanically mixed with the iron. In rolling the blooms into bars or plates the slag particles are elongated into fibers which give the metal its characteristic directional properties.

The products of the first rolling operations may be bundled and heated for rerolling. The high rolling temperatures used and the

Fig. 82. — Longitudinal Section of a Hand Puddled Wrought Iron. (Etched with 5% picric acid in alcohol, often referred to as picral.) X100.

excellent welding characteristics imparted by the slag result in perfect fusion of the several bars or plates in the bundle. Rerolled or double refined plates or pipe skelp are superior to single refined products and command a higher price.[2]

The iron matrix of good quality wrought iron is low in sulphur, manganese, silicon, and carbon. Wrought iron may contain 0.16 to 0.20% silicon, but most of it goes into the slag. About one-half of the total phosphorus content of 0.08 to 0.16% also goes into the slag and the remainder dissolves in the solid iron (ferrite), giving strength and hardness to the alloy, and contributing to its corrosion resistance. Larger amounts, 0.20 to 0.25% phosphorus, would further harden the metal but reduce the ductility and toughness.

The tensile properties of wrought iron, as tested in the longitudinal direction, are as follows[2]:

Tensile strength, lb./sq. in.	42 000–52 000
Yield point, lb./sq. in.	26 000–35 000
Elongation in 8 in., %	40–25
Reduction in area, %	55–40
Brinell hardness	105

Alloying elements, especially nickel, may be used to obtain higher strength wrought irons for special purposes.

The microstructure of a section removed from an old wrought iron bridge is shown in Fig. 82. It shows microscopic variations in grain size and slag content resulting from the fusion, in the rerolling of hand puddled iron, of bars having different characteristics.

In general, the slag content and the mechanical and chemical properties of Aston process mechanically mixed wrought irons are equivalent to those of good hand puddled iron, while the structural uniformity is superior because comparatively large blooms can be made from a single heat by this process.

Transformations in Pure Iron. — The principal complications of the Fe-Fe$_3$C equilibrium diagram are directly associated with the allotropic changes of pure iron. The transformations of iron upon cooling from the molten state to room temperature are indicated by means of a cooling curve in Fig. 83. (The length of the constant temperature transformation interval has been drawn proportional to the heat evolved at each transformation.) The solid phase which freezes from the melt at 1535° C. has a body-centered cubic (B.C.C.) crystal structure and is called delta iron. At 1400° C. the crystal structure of the iron changes from B.C.C. to F.C.C., and at 910° C. it reverts to the B.C.C. form, each transformation
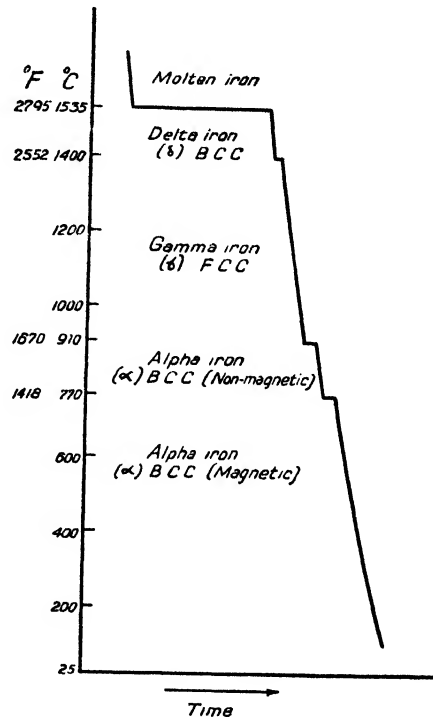
FIG. 83. — COOLING CURVE FOR PURE IRON.

being accompanied by an evolution of heat. The B.C.C. form which is stable below 910° C. is known as alpha iron. For many years the symbol " beta " was applied to non-magnetic B.C.C. iron, existing between 910° and 770° C., in order to distinguish it from the magnetic form existing below 770° C. Although the transformation from non-magnetic to magnetic alpha iron actually occurs over a. range of temperatures, it is accompanied by a marked evolution of heat at about 770° C., which is taken as the transformation temperature. No change in the form or dimensions of the crystal lattice has

been observed, therefore the symbol " alpha " is now used to denote both the magnetic and non-magnetic B.C.C. irons. The heat effect at the magnetic transformation temperature is believed to be caused by an energy change within the individual atoms rather than by re-arrangement of the atoms in the crystal lattice.

The Iron-Iron Carbide Equilibrium Diagram. — The equilib-rium relationships between iron and its carbide, Fe_3C, are repre-
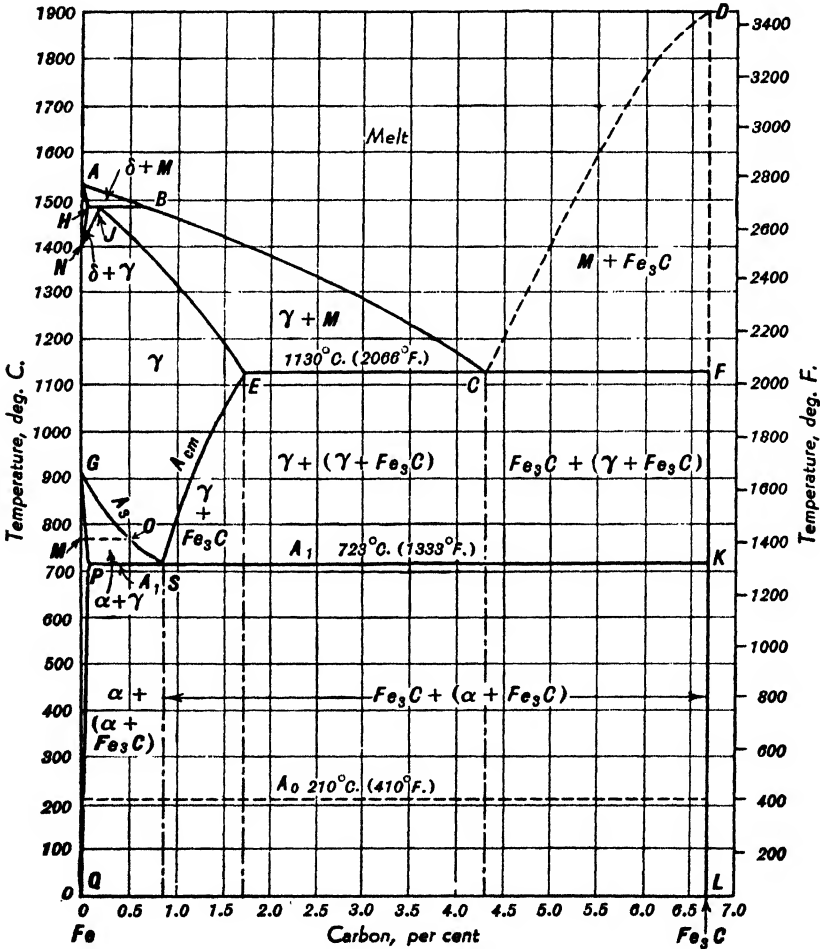
FIG. 84. — THE IRON-IRON CARBIDE EQUILIBRIUM DIAGRAM. (*Courtesy of Samuel Epstein, The Alloys of Iron and Carbon, Vol. I — Constitution, The Engineering Foundation, McGraw-Hill Book Company, New York, 1936.*)

CONSTITUENTS OF THE IRON–IRON CARBIDE SYSTEM

Symbol	Name	Description
α	Ferrite	Solid solution of Fe_3C (or atomic carbon) in B.C.C. or alpha iron
γ	Austenite	Solid solution of Fe_3C (or atomic carbon) in F.C.C. or gamma iron
δ	Delta	Essentially the same as ferrite
Fe_3C	Cementite or iron carbide	Intermetallic compound
$(\alpha + Fe_3C)$	Eutectoid or pearlite	Mechanical mixture of eutectoidal ferrite and Fe_3C.
$(\gamma + Fe_3C)$	Eutectic or ledeburite	Mechanical mixture of austenite and Fe_3C.

sented in Fig. 84. The compound Fe_3C is quite stable in steels, but under certain conditions it may be decomposed to iron and graphite. In such cases the true equilibria are no longer between iron and iron carbide, and the so-called "stable" equilibrium diagram of iron and carbon (graphite) should be used.[3] However, Fe_3C is not decomposed in ordinary steels, hence the "metastable" Fe-Fe_3C diagram is of great value in studying their microstructures and heat treatments.

It is evident from Fig. 84 that increasing carbon contents first increase and then decrease the temperature range of the F.C.C. or austenitic phase. Later we will find that nickel extends this temperature range and stabilizes the austenitic phase to such an extent that certain iron-nickel alloys are F.C.C., or austenitic, even at room temperature. On the other hand, chromium contracts the austenitic range and finally eliminates it entirely at about 12% chromium. It is significant that the crystal structure of nickel itself is F.C.C. while that of chromium is B.C.C. These considerations will be found to be highly important in studying stainless and other alloy steels.

Carbon strengthens and hardens slowly cooled iron-carbon alloys through the formation of the eutectoid $(\alpha + Fe_3C)$. Carbon is responsible for a much greater hardening effect upon heat treatment.

The equilibrium diagram enables us to study the origin and structural nature of the slowly cooled alloys, and provides a starting point for considering the heat treatment of steel. Neglecting, for the present, the complications in the region of the delta phase, the upper portion of the diagram above 1000° C. may be compared with Fig. 26, page 39. If the small beta solution region of Fig. 26 is neglected, or considered equivalent to Fe_3C, the diagrams are quite similar and the fundamental eutectiferous nature of the Fe-Fe_3C diagram becomes obvious. The eutectic mixture, ($\delta + Fe_3C$), solidifies at 1130° C. from molten metal containing 4.3% carbon. The eutectic should contain:

$$\frac{100(4.3 - 1.7)}{6.67 - 1.7} = 52\% \; Fe_3C \; (\text{and } 48\% \text{ austenite})$$

Austenite is a solid solution of carbon in F.C.C. iron. According to some authorities the carbon dissolves as the carbide, Fe_3C, while others believe that carbon atoms are dispersed throughout the lattice structure of the iron, not replacing iron atoms in the pattern, but occupying special positions between the iron atoms. The solubility of carbon in austenite increases from 0.83% to a maximum of 1.7% as the temperature increases from 723° to 1130° C. Practically all steels are welded, rolled, forged, and heat treated while in the austenitic state.

The portion of the Fe-Fe_3C diagram below 1000° C. also resembles an ordinary eutectic system. The solid solution austenite, instead of the usual molten solution, is the phase from which the eutectoid mixture ($\alpha + Fe_3C$) is formed. It will be recalled that the nonferrous eutectoids of Figs. 68 and 76 also form from solid solutions.

The solubility of carbon in alpha or B.C.C. iron is very much lower than in gamma iron. It is approximately 0.04% at the eutectoid temperature and drops to less than 0.01% at room temperature. (Various investigators give 0.025 to 0.043% carbon at the eutectoid temperature and 0.005 to 0.008% at room temperature.)

The Critical Transformation Range. — The region GSP of Fig. 84, known as the critical range, is highly important in the heat treat-
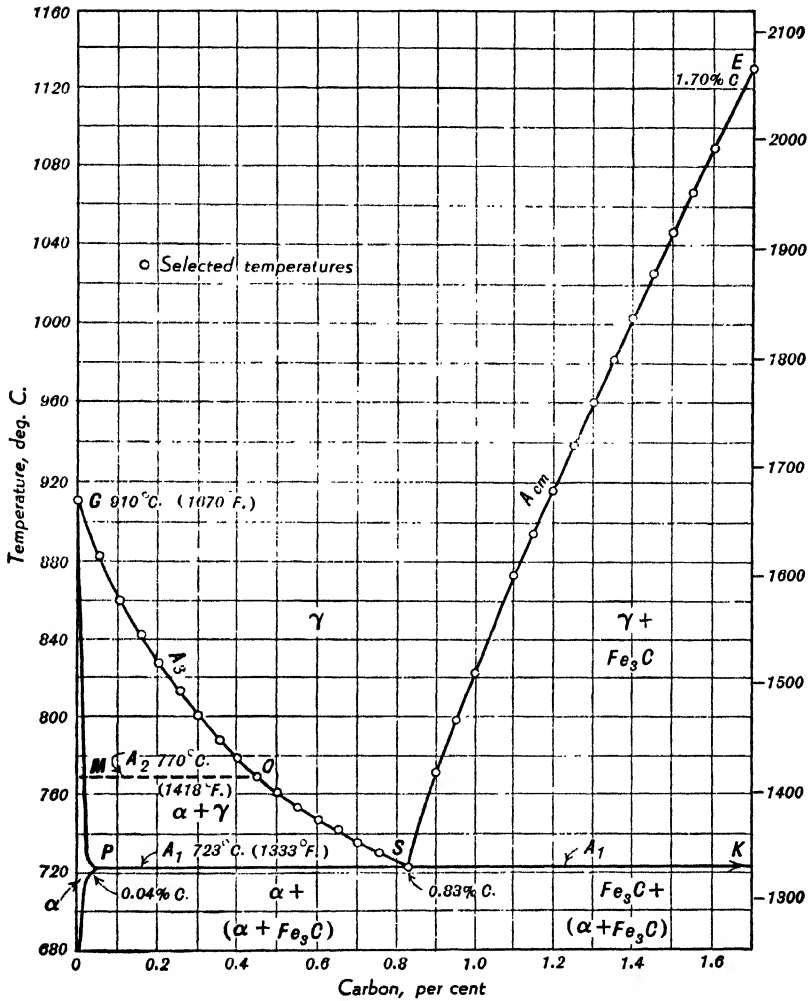
FIG. 85. — THE CRITICAL TRANSFORMATION REGION OF THE IRON-IRON CARBIDE EQUILIBRIUM DIAGRAM. (*Courtesy of Samuel Epstein, The Alloys of Iron and Carbon, Vol. I — Constitution, The Engineering Foundation, McGraw-Hill Book Company, New York, 1936.*)

ment of steels. An enlarged section of this portion of the diagram is shown in Fig. 85. The transformation temperatures will be raised somewhat upon rapid heating and may be lowered by several hundred degrees by rapid cooling, as in water quenching. (See

pages 165 and 207.) The transformation temperatures given in Fig. 85 apply only to very slow heating and cooling conditions.

The letters " r " and " c " are often used to distinguish between transformations upon cooling and heating, respectively, as indicated in the following list of definitions.*

DEFINITIONS OF THE CRITICAL POINTS OF STEELS

TRANSFORMATIONS IN HYPOEUTECTOID STEELS (LESS THAN 0.83% CARBON)

Ar_3 — Beginning of the precipitation of ferrite from austenite on cooling.

Ar_2 — Transformation of non-magnetic ferrite to magnetic ferrite on cooling.

Ar_1 — End of the precipitation of ferrite from austenite on cooling. All remaining untransformed austenite changes completely to pearlite.

Ac_1 — All pearlite transforms to austenite on heating. Beginning of the solution of ferrite in austenite.

Ac_2 — Transformation of magnetic ferrite to non-magnetic ferrite on heating.

Ac_3 — End of the solution of ferrite in austenite on heating.

Ar' — This symbol is used when the transformation of austenite to pearlite takes place at a temperature lower than the normal Ar_1 temperature. (See page 207.)

Ar'' — This symbol is used when the cooling rate exceeds a certain critical rate and the austenite transforms to martensite. (See page 207.)

TRANSFORMATIONS IN HYPEREUTECTOID STEELS (MORE THAN 0.83% CARBON)

Acm — Beginning of the precipitation of cementite from austenite on cooling.

Ar_1 — End of the precipitation of cementite from austenite on cooling. All remaining austenite transforms to pearlite.

Ac_1 — Transformation of all pearlite to austenite on heating. Beginning of the solution of excess cementite in austenite.

Acm — End of the solution of excess cementite in austenite on heating.

A_0 — Cementite loses its magnetism upon heating above this temperature. (Both hypoeutectoid and hypereutectoid steels)

* Ar = arrest on cooling — " r " is from the French " refroidissement," meaning cooling.

Ac = arrest on heating — " c " is from the French " chauffage," meaning heating.

Transformations of a Hypoeutectoid Steel upon Cooling through the Critical Temperature Range. — A rolled bar of 0.2% carbon steel will have a homogeneous austenitic structure upon heating to 1000° C. (1832° F.). Upon cooling very slowly certain changes take place in the microstructure which are accompanied by thermal, magnetic, and dimensional changes. The time-temperature cooling curve of Fig. 86, representing ideal conditions obtained upon extremely slow cooling, will be convenient as a basis for studying these changes. (See also Fig. 85.)

The gradual separation or precipitation of ferrite from austenite between Ar_3 and Ar_1 is accompanied by an evolution of heat which delays cooling, as shown by the departure from the dotted line representing the natural cooling rate assuming no transformation takes place. The magnetic change at Ar_2 also occurs over a range of temperatures but there is a decided maximum at 770° C. and

FIG. 86. — COOLING CURVE FOR A 0.20% CARBON STEEL COOLED VERY SLOWLY THROUGH THE CRITICAL TEMPERATURE RANGE.

the heat evolved is sufficient to prevent lowering of the temperature for an appreciable period of time. It is interesting to note that no dimensional change occurs at this temperature which is additional evidence in support of the belief that Ar_2 does not represent a phase change. At Ar_1 the formation of pearlite from austenite at constant temperature causes a final delay in cooling.

Very slow heating produces the reverse transformations at the same temperatures.

Practical cooling and heating rates give transformations which deviate from those given in the equilibrium diagram as shown by Mehl and Wells' data, Fig. 87, taken at cooling and heating rates of ½° C. per minute. (Note that Mehl and Wells place the eutectoid composition at 0.80% carbon.) In the heat treatment

of steel the temperatures used for hardening or annealing are generally about 25° C. (50° F.) or more above the critical temperature in order to insure complete transformation.

The Ar_1 or eutectoid transformation will be better understood after studying the conditions prevailing just before and just after the change. At 725° C. (slightly above Ar_1) the proportion of primary ferrite which has formed in the 0.20% carbon steel will be:

$$\frac{(0.83 - 0.20)100}{0.83 - 0.04} = 80\% \text{ ferrite}(20\% \text{ untransformed austenite}).$$

At 720° C. the remaining austenite will have transformed to pearlite (eutectoid), giving 80% ferrite-20% pearlite. At room temperature the solubility of carbon in ferrite is taken as zero for all practical purposes, hence the calculation for ferrite is slightly altered, giving 76% ferrite-24% pearlite.

It is important to note that pearlite itself contains a considerable proportion of ferrite. (Eutectoidal ferrite — see Fig. 92.) At room temperature pearlite contains:

$$\frac{(6.67 - 0.83)100}{6.67 - 0} = 87.5\% \text{ eutectoidal ferrite } (12.5\% \text{ Fe}_3\text{C}).$$

The total ferrite (primary plus eutectoidal) in the 0.20% carbon steel is readily calculated as:

$$\frac{(6.67 - 0.20)100}{6.67 - 0} = 97.0\% \text{ total ferrite } (3.0\% \text{ Fe}_3\text{C}).$$

The transformations occurring in hypereutectoid steels will be considered in Chapter XV.

Microstructures of Slowly Cooled Steels. — The microstructures of slowly cooled steels yield useful information on grain size, carbon content, structural uniformity, slag or inclusion content, and on certain related factors such as hardenability.

The microstructure of an annealed wrought steel containing 0.15% carbon is shown in Figs. 88 and 89. The white areas are the primary ferrite which formed from austenite between Ar_3 and Ar_1. The black areas are pearlite (eutectoid) which formed from the remaining austenite at Ar_1. Practically all the carbon is present as Fe_3C plates in the pearlite. This structure, called laminated

pearlite, is characteristic of the eutectoid after cooling slowly from the austenitic state.

The microstructure of a 0.60% carbon steel, containing approximately ¾ pearlite and ¼ ferrite, is illustrated in Fig. 90.

The microstructure of pearlite as it occurs in eutectoidal steels which have been cooled slowly from above the A_1 critical temperature is shown in Figs. 91–93. At low magnification the entire structure appears dark but non-uniform in shading from grain to grain. The etched surface has a pearly appearance when viewed in natural light without magnification. Higher magnifications resolve the well laminated structure of the individual pearlite grains as in Figs. 92 and 93. In the latter photomicrograph the cementite plates in the large grain of pearlite appear to be embedded

Fig. 87. — The Eutectoid Region of the Iron-Iron Carbide Diagram Showing Degree of Undercooling and Superheating of the Transformations at Rates of Cooling and Heating of ½° C. per Minute. (*Courtesy of Robert F. Mehl and Cyril Wells, Trans. Am. Inst. Mining and Met. Eng.*[4])

in the ferrite, but to project from the surface about 3/16 inch at X1400, giving a structure having the appearance of fish scales. Apparently the cementite plates were cut at an angle in polishing and the etching reagent dissolved the ferrite surrounding the cementite, leaving the plates projecting from the new level. Cementite is very resistant to attack by picral and nital, the etching reagents ordinarily used. Ferrite is readily dissolved and may appear to be darkened by prolonged etching with nital. It is apparent from Fig. 93, however, that both phases are light in color when fully resolved even after deep etching. The dark appearance of pearlite at low magnification is caused by poor resolution of the structure and by the

shadows cast by the cementite particles which stand in relief after etching.

The laminations may be of such a fineness that they are imperfectly resolved even at high magnification as illustrated by Fig. 94, the structure of a hot rolled eutectoid steel cooled in air from the
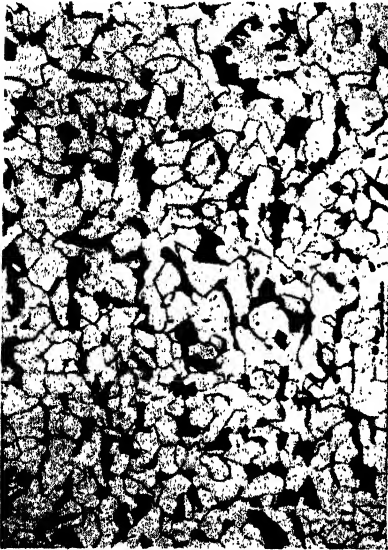
Left. FIG. 88. — 0.15% CARBON STEEL. COOLED SLOWLY FROM 1650° F. (Etched with 3% nital.) Brinell 120. X100.

Right. FIG. 89. — SAME AS FIG. 88. X1000.

finishing temperature. Rapid cooling and alloying greatly alter the appearance of pearlite, giving finer lamellar structures.

Hypereutectoid steels which have been cooled slowly from the gamma region have excess cementite in addition to pearlite in their microstructures as shown in Fig. 95. The special etching reagent (alkaline sodium picrate) used for Fig. 96 darkens the cementite, especially the massive cementite network in the grain boundaries.[5] In this way ferrite and cementite may be definitely distinguished since massive ferrite areas are uncolored after proper application of the sodium picrate etch. Upon considering tool steels and carburized steels in more detail it will become apparent that

network structures such as represented by Figs. 95 and 96 are to be avoided in commercial products because of their brittleness.

Effect of Carbon on Annealed Steels. — The microstructural evidence presented thus far should acquaint the reader with the constituents present in slowly cooled carbon steels; namely, ferrite, laminated pearlite, and cementite. It is apparent that the microstructures depend to a large extent on the carbon content. The effect of carbon upon the mechanical properties is equally marked. In annealed steels carbon in the form of cementite has a direct hardening effect. Cementite itself is very hard and brittle but in the form of laminated pearlite it increases the strength as well as the hardness. The tensile strength in the annealed state varies from about 40,000 lb. per sq. in. for pure iron to 115,000 lb. per sq. in. for eutectoidal steels. The corresponding ductilities are approximately 40 to 10% elongation in 2 inches. Impact toughness decreases rapidly with carbon content in this range. Further in-

FIG. 90. — 0.60% CARBON STEEL. COOLED SLOWLY FROM 1550° F. (Etched with 3% nital.) Brinell 183. X100.

crease in carbon content over 0.83% increases the hardness but has little additional effect on the tensile strength or ductility provided structures having cementite networks, as shown in Fig. 95, are avoided.

In steels hardened by heat treatment carbon is even more effective in increasing the hardness and strength.

Effect of Other Elements. — The elements sulphur, phosphorus, silicon, and manganese, when present in the usual amounts, influence the structure and properties of steels to a much smaller extent than does carbon. However, each element adds certain character-

Left. FIG. 91. — EUTECTOIDAL STEEL CONTAINING APPROXIMATELY 0.75% CARBON. (Etched with picral.) Rockwell "B" 100. X100.

Right. FIG. 92. — EUTECTOIDAL STEEL. (Etched with picral.) X500.

Left. FIG. 93. — SAME AS FIG. 92. (Deeply etched.) X1400.

Right. FIG. 94. — EUTECTOIDAL STEEL CONTAINING APPROXIMATELY 0.8% CARBON. (Etched with picral.) Rockwell "C" 25. X1500.

istic qualities, and each must be properly controlled to insure suitability of the product for its ultimate application.

Sulphur. — It is difficult to class any one of these elements as an outright impurity. Even sulphur functions as a useful alloying element in certain free-cutting steels to be mentioned later. However, the principal action of sulphur is its interference with hot rolling
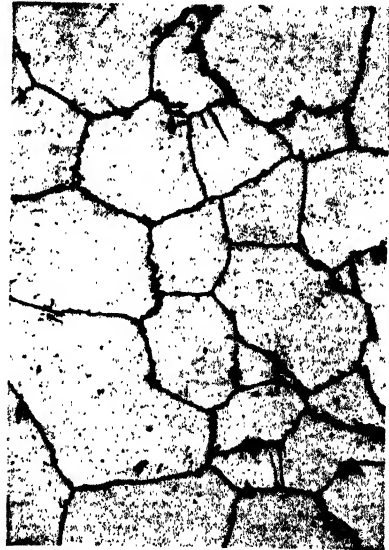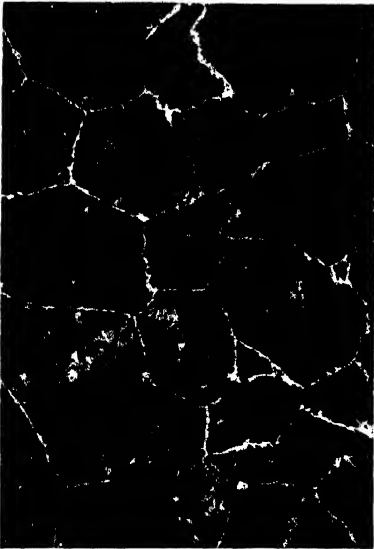
Left. FIG. 95. — HYPEREUTECTOID STEEL CONTAINING APPROXIMATELY 1.2% CARBON. (Etched with picral.) Rockwell "C" 24. X100.

Right. FIG. 96. — SAME AS FIG. 95. (Etched by electrolysis in alkaline sodium picrate.) X100.

and forging operations. Sulphur exists in carbon steel as FeS and MnS, the former having a strong tendency to form films and small particles at the grain boundaries. These FeS inclusions soften and may actually melt at rolling and forging temperatures, reducing the tenacity of the hot metal and sometimes causing disintegration by cracking in the rolls or under the hammer. Fortunately manganese has a strong affinity for sulphur, and MnS inclusions, which are more refractory than FeS, are not as likely to segregate at grain boundaries and are sufficiently plastic to deform with the metal during forging or rolling. The manganese content should always

be high enough to combine with the sulphur in steels to be hot worked. Recommended ratios of manganese to sulphur range from 3 to 1 to about 8 to 1, depending on the type of iron or steel and the manner in which it is to be processed. When this ratio is low the metal may have a " critical " temperature range (not to be confused with the Ac and Ar transformation temperatures) in which hot working should not be done because of the danger of cracking. This is known as the red-short range.

The sulphur content of ordinary open hearth steels is usually under 0.05%. Electric furnace steels and certain open hearth grades often have sulphur contents of 0.025% and under.*

Sauveur, in " The Metallography and Heat Treatment of Iron and Steel " [5], has an extended discussion of the effect of sulphur on steel. This book, which was first published in 1916, is a treatise on iron and steel and may be used as a general reference.

Phosphorus. — Unlike sulphur, phosphorus does not ordinarily form slag inclusions in steel, but dissolves in the ferrite, increasing the strength and hardness and improving the resistance to corrosion.

Phosphorus has been held responsible for many types of brittleness, including " cold-shortness " or reduction in impact toughness at low temperatures. Embrittlement by phosphorus may be due in part to its tendency to coarsen grain size, as well as to harden the ferrite; however, the effect of phosphorus on toughness is usually not serious except in the higher carbon steels. There is little doubt but that phosphorus aids considerably in strengthening wrought iron and certain low carbon steels which contain up to 0.20% phosphorus as a definite part of their specified compositions.[6] On the other hand, it is equally well known that the embrittling effect of the higher phosphorus contents is a definite limitation to the amount of this element which may be used as an alloying addition for making high strength steels.

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* Relative sulphur content and its distribution may be determined by making a sulphur print. Photographic paper is moistened in very dilute H_2SO_4 and pressed against the polished or ground surface of the section to be tested. Soluble sulphides in the metal leave a brown stain on the paper. The print may be preserved by " fixing " in the usual way.

Phosphorus has a tendency to segregate in steels, both on a microscopic and a macroscopic * basis, which increases the possibility of embrittlement.

Phosphorus content is subject to control in the basic open hearth and in electric furnaces but not in the acid Bessemer converter. It is usually held to about the same limits as sulphur and may be reduced to very low contents in certain basic open hearth grades.

Silicon. — Silicon and manganese often assume the roles of important alloying elements in steels and are nearly always present, in smaller amounts, in plain carbon steels. Both elements are also effective deoxidizers and manganese is almost indispensable for large scale steel making operations.

Silicon, like phosphorus, dissolves in the ferrite, increasing its strength and hardness without lowering the ductility appreciably unless present in larger amounts than are ordinarily used in plain carbon steels. The silicon content ordinarily varies between 0.10 and 0.35% except in " rimming " steels which contain only a trace of silicon.

Silicon contents of 0.3 to 0.5% have a beneficial effect on steel castings, increasing the strength without loss of ductility, probably because of the greater soundness of the castings made possible by the degasifying and cleansing action of silicon.

Manganese. — Manganese is a much more potent addition than silicon in many respects. Its beneficial action in controlling the form of the sulphur and its value as a mild deoxidizer have already been mentioned.

Manganese not combined with sulphur or oxygen in solid steel is free to act in two ways. If the carbon content is low, most of the manganese will dissolve in the ferrite, adding to its strength and hardness. With higher carbon content, it is believed that Mn_3C forms and becomes associated with Fe_3C, producing even greater increase in strength and hardness at some sacrifice of ductility. According to H. H. Campbell the strengthening effect does not become evident until over 0.30% manganese is present, and the actual

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* Structures observed at magnifications less than X10 are called macrostructures. The general appearance to the naked eye of a section of an ingot, casting, or forging may also be referred to as its macrostructure.

increase varies with the carbon content, as shown in the following data reported by Stoughton.[7] The second column gives the increase in tensile strength of basic open hearth steel as produced by each 0.01% manganese, beginning with 0.30% manganese.

Carbon content of the steel, %	Increase in tensile strength for each 0.01% Mn over 0.30% Mn lb./sq. in.
0.05	110
0.10	130
0.15	150
0.20	170
0.25	190
0.30	210
0.35	230
0.40	250

Although the additional effects of heat treatment have not been introduced, it may be mentioned at this time that the relationship between the manganese and carbon contents controls, to a large extent, the critical cooling rate and consequently the quenching medium to be used for hardening.

Manganese and other alloying elements such as chromium and nickel reduce the carbon content of the eutectoid. The pearlite present in annealed steels may thus contain less than 0.83% carbon, depending on the composition of the steel.

The equilibrium relations of the Fe-Mn-C alloys have been studied in detail. Walters and Wells [8] have summarized this work.

Banding in Steel. — Phosphorus, silicon, and certain other elements sometimes present in steels, together with the MnS and other slag inclusions, tend to become segregated in the microstructures of castings or ingots. Even after rolling and forging, these elements sometimes remain segregated, the inclusions being deformed and elongated in the direction of hot working.[3] Carbon is known to be less soluble in these areas than in the regions of lower impurity content, which adds further to the segregated condition of the microstructure, giving distinctly banded structures as shown in Fig. 97. The black bands are pearlite while the gray stringers are slag inclusions. This structure is characteristic of high sulphur, free-machin-

ing steels; however, carbon and especially alloy steels with normal sulphur contents are also subject to banding. Banding gives steel directional properties, the impact toughness and ductility being reduced in the transverse direction, which is usually considered to be detrimental in high strength steel for gears, tubing, and forgings, as well as for sheet and plate which must withstand forming operations.

Oxygen and Nitrogen in Steel. — Oxygen is present in steel as iron and manganese oxide and silicate inclusions, and to a lesser extent as oxygen dissolved in the ferrite. The type and amount of inclusions present, hence their influence on the properties of the steel, will depend upon the deoxidizers used and other details of the steel making process which are not discussed in this textbook but may be found in " The Physical Chemistry of Steel Making "[9] and " The Alloys of Iron and Carbon."[3]

Briefly it may be stated that small, rounded oxide inclusions,

Fig. 97. — S.A.E. 1112 Steel Showing Banding. (See page 189.) (Etched with picral.) X100.

evenly distributed, are less detrimental to mechanical properties than are elongated or stringer types which reduce the transverse ductility and impact toughness. In extreme cases resulting from poor steel making practice, segregated stringer type inclusions may weaken a sheet to such an extent that separation or splitting occurs in rolling or in fabricating operations. Generally the oxides are of the rounded type while sulphides and silicates tend to elongate in the direction of rolling or forging. It is believed that very small visible and submicroscopic oxide inclusions, principally Al_2O_3, are responsible for inhibiting grain growth in certain types of fine grained, killed steels.

Larger inclusions are known to have an adverse effect on impact

and *fatigue resistance* because their unyielding nature tends to break the continuity of stress distribution, resulting in local areas of high stress concentration about each inclusion. Soft ferritic steels are capable of redistributing the stress through local plastic defor-

TABLE XXII.   DISTRIBUTION OF FINISHED ROLLED STEEL
BY PRINCIPAL CONSUMING GROUPS

	1937	1929	1923
Automotive	20.04%	17.57%	10.59%
Buildings	12.32	14.70	13.53
Railroads	11.74	18.44	31.02
Containers	8.71	4.67	3.68
Exports	6.74	4.83	6.22
Oil, gas, water	5.84	9.01	10.60
Machinery	4.88	3.81	2.63
Agriculture	3.66	5.27	2.34
All other	12.79	10.65	19.39
	86.72	88.95	
Jobbers	13.28	11.05	
	100.00%	100.00%	100.00%

BY PRODUCTS

		1937	1929
Galvanized sheets	4.0		
Black plate for tinning	8.2		
All other black plate and sheets except galvanized sheets and tin plate	19.4	40.7%	28.7%
Strip steel *	9.1		
Bars and shapes under 3 inches	14.4	16.9	20.2
Concrete bars	2.5		
Plates		9.3	11.5
Structural shapes over 3 inches		7.5	10.8
Rails		4.4	7.4
Track accessories		1.9	2.9
Tubing and pipe		10.2	10.8
Wire products		6.6	6.0
All other finished steel products		2.5	1.7
		100.0%	100.0%

* Up to 24 inches wide. 1938 size limitations for strip are given on page 176.

mation, thus reducing the stress-raising effect of inclusions to a greater extent than in high strength heat treated steels.

When the nitrogen retained upon solidification of steel surpasses the solubility limit, which is very low at room temperature [3], it forms needles of iron nitride or possibly other compounds such as aluminum nitride. Since the solubility of these compounds changes with temperature, precipitation hardening effects are possible.

The remarkable hardening effect of high nitrogen contents in case hardening steels is discussed in Chapter X.

Distribution of Rolled Steel Products. — The data in Table XXII showing the distribution of finished steel by products and by principal consuming groups are reproduced through the courtesy of " Steel." [10]

The total production of finished steel was 32,878,000 gross tons in 1937. Semifinished steel accounted for an additional five million gross tons.

Estimates of the total tonnage of alloy steel produced vary because of the difficulty in obtaining complete data. A production of somewhat less than one million gross tons of alloy steels was reported to " Steel " in 1937, approximately 60% of this amount being used in the automotive industry. The total estimated production of alloy steels is much greater than this, approximately 3,200,000 gross tons of ingots and castings.

Flat Rolled Steel. — The classification given in Table XXIII is now used by the Association of American Steel Manufacturers Technical Committees to differentiate between the various flat rolled steel products.

Bars and plates are usually hot rolled products used for machine or general construction purposes. Certain types of bars may be cold finished by rolling or drawing.

Both hot rolled and cold rolled strips and sheets may be supplied in coils as well as cut lengths, and according to the above schedule the principal distinctions between sheet and strip are based on width and thickness rather than length.

The terminology of the steel industry is very much influenced by tradition, and many seemingly arbitrary terms and classifications are in use. The standards for flat rolled steel are a typical ex-

ample.   Authoritative information on present practice will be found
in a series of " Steel Products Manuals " published by the Ameri-
can Iron and Steel Institute, beginning in 1937.

From the standpoint of surface finishes of sheets the principal
classifications are: hot rolled or black, hot rolled and pickled, and
cold rolled bright annealed.   Hot rolled finishes are used for tanks,
smoke stacks, pipe, and other products which do not require a high
finish and are not drawn or subjected to severe forming operations.
The scale on hot rolled steel scores drawing and forming dies, there-
fore a hot rolled and pickled finish is used for parts which do not
require a high finish but are subject to deformation in fabrication.

TABLE XXIII.   STANDARD CLASSIFICATION BY SIZE OF FLAT-
ROLLED CARBON-STEEL   **Hot Rolled**

Widths, in.	Thicknesses, in.						
	.2500 or thicker	.2499 to .1875	.1874 to .0568	.0567 to .0344	.0343 to .0255	.0254 to .0142	.0141 or thinner
Up to 3½ incl.......	Bar	Strip	Strip	Strip	Strip	Sheet	Sheet
Over 3½ to 6 "	Bar	Strip	Strip	Strip	Sheet	Sheet	Sheet
" 6 to 12 "	Plate	Strip	Strip	Sheet	Sheet	Sheet	Sheet
" 12 to 32 "	Plate	Sheet	Sheet	Sheet	Sheet	Sheet	T. M. Black
" 32 to 48 "	Plate	Sheet	Sheet	Sheet	Sheet	Sheet	Sheet
" 48	Plate	Plate	Sheet	Sheet	Sheet	Sheet	——

Cold Rolled

Widths, in.	Thicknesses, in.		
	.2500 or thicker	.2499 to .0142	.0141 or thinner
Up to 12 incl..............	Bar	Strip	Strip
Over 12 to 24 "	Strip *	Strip *	Strip *
" 12 to 24 "	Sheet **	Sheet **	T. M. Black **
" 24 to 32 "	Sheet	Sheet	T. M. Black
" 32	Sheet	Sheet	Sheet

* When a definite temper as defined in A.S.T.M. specification A–109, or a special
edge, or special finish is specified.

** When no special temper, edge, or finish is specified.   (T.M. = Tin Mill)

Frames and other unexposed parts of automobiles are made of hot rolled and pickled steel. Cold rolled bright annealed finishes are applied to sheets for automobile outside body parts, metal furniture, and other products requiring a high quality paint or lacquer finish.

Sheet Steel. — The outstanding metallurgical development of recent years is the great increase in production and improvement in quality of sheets. The data in Table XXII show that the relative production of sheet and strip increased from about 29% of the total production in 1929 to about 41% in 1937, while the production of heavier structural products such as rails, shapes, and plates decreased. The increased production of thin gage, large size, high quality sheets for automobiles, furniture, refrigerators, and countless other stamped, formed, and welded products has been made possible by the development of continuous sheet and strip rolling mills, some of which are capable of producing high finish sheet in widths up to 98 inches.

Considering steel consumption by industries, the correlation between increased use of steel by the automotive industry and increased production of sheet and strip is notable. Although railway construction has been decreasing for several years, the development of higher strength, corrosion resisting steel has made it possible to design lighter weight rolling stock, thereby increasing the relative proportion of light gage plate and sheet metal used in this field.

Tin plate for food containers remains one of the stable, high tonnage products utilizing very low carbon steel sheets. The use of sheet metal containers for other purposes is increasing.

Also included among the long established applications for low carbon steel are a wide variety of black, galvanized, and terne coated sheets (see Chapter X) for building purposes, including siding, roofing, gutters, heating and ventilating ducts, stacks, and concrete forms.

The base price of heavy gage hot rolled sheets is about $43 per ton (January, 1939). Sheet prices are subject to extras or deductions for gage, size, and surface and subject to extras for quantity, flatness, gage tolerance, crating, and oiling.

Cold rolled deep drawing 20 gage fender stock was about $62 per ton in 1938 compared with $125 per ton in 1923. This reduc-

tion in the price of sheets and the remarkable improvement in the quality of many grades, are the result of manufacturing economies made possible by the continuous rolling process and by research in all phases of production.

Stamping and Drawing Sheets. — As already indicated, a large part of the increased production of sheets is in stamping or deep drawing stock for press operations. The entire metallurgy, from melting practice to heat treatment and the final finishing operations, has been specialized to enhance the plasticity of the metal and the quality of the surface of these sheets.[11, 12, 13, 14]

FIG. 98. — OPEN HEARTH STEEL FOR DEEP DRAWING PURPOSES, CONTAINING 0.06% CARBON. (Etched with 3% nital.) Rockwell " B " 40. X100.

Sheets for deep drawing or stamping are ordinarily made of very low carbon open hearth steel (0.04 to 0.10% carbon) finished and cast into ingots by the " rimming " process.[15, 16, 17] The surface of these ingots has a lower carbon and impurity content than the interior, and the sound surface layer, which may be several inches thick on a large ingot, is responsible for the relative freedom from seams, scabs, and other defects on the surfaces of the finished sheets. Rimming steel ingots are also characterized by internal blowholes which constitute from 2 to 4% of the total volume and are the result of the evolution of carbon-monoxide gas during the solidification of the ingot. These blowholes are effectively welded during hot rolling under normal conditions.

In continuous mill practice the ingots are hot rolled to coiled " thin bar " approximately 0.1 inch in thickness and from about 24 to nearly 100 inches in width. These coils may be cold reduced to finish gage in a continuous mill, or cut to lengths, reheated, and hot rolled in packs to finish gage. In either case certain pickling,

normalizing, annealing, temper rolling, roller leveling, and shearing operations are necessary to produce a finished sheet. The number of operations required is materially less for cold reduction than for hot reduction processing cycles, but the cost of cold reduced sheets is higher because of more rigid inspection standards and closer gage control. The cold reduction process offers many advantages in surface quality, mechanical properties, economy, and flexibility of operation over the hot reduction method.[14]
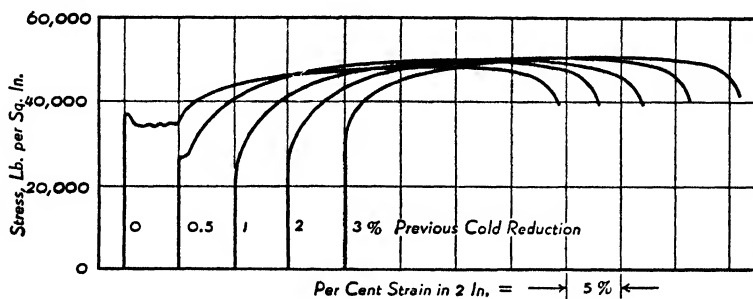
Fig. 99. — Stress-Strain Curves of Mild Steel Sheets Subjected to the Indicated Amounts of Cold Work prior to Tensile Testing. (*Courtesy of R. O. Griffis, R. L. Kenyon, and R. S. Burns,*[11] *Trans. Am. Soc. for Metals.*[12])

Both hot and cold reduced sheets require a final heat treatment to develop high ductility for press work. Hot reduced sheets are generally normalized (air cooled from above the Ac_3 critical temperature), pickled, and bright annealed at a temperature below the critical range to relieve normalizing stresses. Cold reduced sheets are cold rolled 40 to 60% and recrystallized by a bright box annealing treatment at a temperature near or below the Ac_1 point.   (See Fig. 14, page 15.)

The purpose of temper rolling and roller leveling, as applied after the final annealing treatment, will be explained in the next section.

A typical microstructure of a cold reduced drawing sheet is shown in Fig. 98. Carbon is present as very small carbide particles in the equi-axed ferrite grains. Elongated grains resulting from incomplete recrystallization in the annealing operation (see Figs. 12 and 13, page 15) give undesirable directional properties. Too fine a grain size may result in a stiff sheet which may break or buckle in

the dies. Too coarse a grain size gives rough surfaces after drawing or forming, as in the case of coarse grained copper alloys.[18]

Mechanical Properties of Sheet Steels for Stampings. — Yield strength, tensile strength, and per cent elongation are not sufficient, in the case of low carbon sheet steels, to define their forming charac-
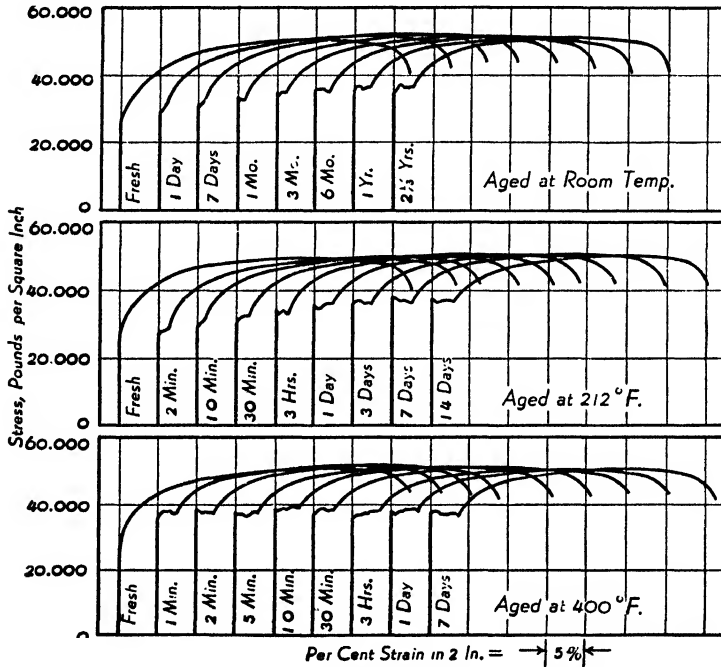
FIG. 100. — STRESS-STRAIN CURVES OF MILD STEEL SHEETS COLD ROLLED 1% AND AGED FOR THE INDICATED TIMES AT THREE DIFFERENT TEMPERATURES. (*Courtesy of R. O. Griffis, R. L. Kenyon, and R. S. Burns,*[11] *Trans. Am. Soc. for Metals.*[12])

teristics. In common with other low carbon steels, drawing sheets are subject to the occurrence of a yield point when tested in tension, as illustrated by several of the stress-strain curves in Figs. 99 and 100. Although this condition is an indication of good ductility and serves the useful purpose of relieving stress concentrations in structural steel, it is undesirable in high finish sheets for press work because of the poor surface obtained after stamping or drawing. The deformation at the yield point, called yield point elongation, results in sudden local stretching at various locations in the sheet. Drastic

stretching is necessarily accompanied by a correspondingly large reduction in thickness, resulting in an unsatisfactory surface as shown in Fig. 101. These strain markings are variously known as "stretcher-strains," "Lüder's lines," "Hartmann's lines," "flow figures," "orange peel," or "worms." The fundamental reason for their occurrence in mild steel and pure iron and only rarely and under special conditions in other metals is still a controversial matter.

The possibility of eliminating the yield point, and the corresponding stretcher-strains, by means of drastic cold rolling is evident in the stress-strain curves of Fig. 10, page 14. However, such treatment reduces the plasticity to such an extent that successful press operations would be impossible. Fortunately, much smaller amounts of cold rolling, of the order of 1% reduction in thickness, are sufficient to eliminate the yield point elongation, as shown in Fig. 99. The ductility is only slightly reduced by this process, known as temper rolling.

FIG. 101. — STRETCHER STRAINS IN A DRAWING SHEET HAVING A HIGH YIELD POINT ELONGATION. (*Courtesy of T. F. Olt, Trans. Am. Soc. of Mech. Eng.*[14])

Alternate bending and reverse bending between the small diameter rolls of a roller leveler also tends to cold work the metal and eliminate the yield point elongation. Temper rolling must always precede roller leveling because soft annealed sheets tend to "break" (yield locally) in the roller leveler. A combination of temper rolling and roller leveling is often applied, the latter operation being of such a nature that the fabricator may put his sheets through the leveler shortly before use in his own plant.

Stress-strain curves for temper rolled sheets are of the rounded type (see Fig. 99), consequently the yield strength is reported

rather than yield point.[*][19,][20] The amount of reduction by temper rolling required to produce a smooth stress-strain curve, indicating freedom from stretcher-strains, will depend upon several factors including the original yield point elongation, grain size, and diameter of the rolls.[13]

TABLE XXIV. EFFECT OF TIME ON THE PHYSICAL PROPERTIES OF MILD STEEL SHEETS TEMPERED 1 PER CENT

Hot-reduced, Normalized, and Annealed Before Temper Rolling

Time	Yield Strength lb./sq. in.	Tensile Strength lb./sq. in.	Elongation in 2 in., %	Yield Point Elongation, %	Rockwell " B " Hardness
Fresh	28 000	49 950	39.9	0.0	47.6
1 day	28 400	50 000	38.9	0.2	48.3
7 days	31 860	50 470	37.7	0.5	49.4
30 days . . .	33 670	50 460	38.3	0.8	50.1
90 days . . .	34 770	50 530	37.0	1.2	—
180 days . .	35 940	50 960	35.2	1.6	—

Cold-reduced and Annealed Before Temper Rolling

Time	Yield Strength lb./sq. in.	Tensile Strength lb./sq. in.	Elongation in 2 in., %	Yield Point Elongation, %	Rockwell " B " Hardness
Fresh	23 770	45 120	42.2	0.0	44.1
1 day	24 450	44 750	41.8	0.4	42.6
7 days	27 600	44 350	41.0	1.5	42.8
30 days . . .	31 210	44 980	40.0	2.5	44.1
90 days . . .	32 030	44 400	39.0	3.4	45.6
180 days . .	32 750	44 900	38.0	3.8	46.0

Courtesy of T. F. Olt, Trans. Am. Soc. of Mech. Eng.[14]

If the sheets are to be fabricated soon after temper rolling, the smallest reduction which will eliminate yield point elongation is naturally preferred. However, any considerable delay in using the sheets may cause return of the yield point by aging, as illustrated by the data in Table XXIV [14] and the room temperature curves of Fig. 100. Heavier reductions by temper rolling delay the reversion to the original state, while elevated temperatures hasten the process. Davenport and Bain [12] have pointed out that the return of the yield point occurs simultaneously with " strain-aging " in the commercial low carbon, oxygen bearing steels. As previously men-

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* The 0.5% yield point, as defined and determined in reference 20, is now called the 0.5% yield strength.

tioned, oxygen is partially soluble in ferrite, hence strain-aging following cold working may be caused by precipitation of an iron-oxygen compound along the slip planes.   Regardless of the mechanism by which the aging phenomenon affects the yield point, it has been shown that deoxidized or killed steels when properly heat treated are not subject to return of the yield point and the stretcher-straining tendency after cold working.[12, 21, 22]   Non-aging drawing steels are now available and further development work in this field is to be expected.

Davenport and Bain [12] have shown that strain-hardening is quite distinct from the precipitation hardening or quench aging effect (which may, however, take place concurrently) and that carbon is principally responsible for quench hardening.   (See Fig. 41, page 73.)

In addition to the drawing and deep drawing grades, harder grades are available for certain products requiring only blanking or stamping operations.   The compositions of the five grades of cold rolled strip steel included in A.S.T.M. Standard A109–38 are essentially the same, the difference in temper being produced by the amount of cold work and other processing variables.

The approximate properties for the various tempers (not specification requirements) are listed in an appendix to Standard A109–38 and are reproduced as Table XXV.

The chemical requirements for grades 1 and 2 are: 0.04 to 0.12% carbon, with up to 0.25% permissible for some uses.   The requirements for grades 3, 4, and 5 are: 0.04 to 0.12% carbon, 0.30 to 0.50% manganese, less than 0.04% phosphorus, and less than 0.055% sulphur.

Other Uses for Low Carbon Steels. — Wire products consume an appreciable part of the total steel production.   A large proportion of this wire is made from low carbon steel for use as galvanized fencing and wire nails.

Low carbon steels containing 0.15 to 0.25% carbon are suitable for applications where moderate strength and hardness, rather than high plasticity are required.   Ordinary constructional or building steels, which are commonly in this range, are usually semi-killed steels melted in open hearth furnaces and teemed into large, big end

TABLE XXV. APPROXIMATE PHYSICAL PROPERTIES FOR VARIOUS TEMPERS OF COLD ROLLED STRIP STEEL

(These values are given as information only and are not intended as criteria for acceptance or rejection.)

Grade or Temper	Hardness (Rockwell "B" Scale), ($\frac{1}{16}$-in. Ball 100-kg. Load)	Depth of Cup for 0.050-in. Thickness of Strip mm.[a]	Tensile Strength, psi.[b]	Elongation in 2 in. for 0.050-in. Thickness of Strip. per cent [c]	Remarks
No. 1. Hard........	90 + 6[d]	6 to 7	80 000 + 12 000	3 ± 2	Intended for flat blanking only.
No. 2. Half-hard	80 ± 5	7 to 8	64 000 ± 8 000	9 ± 5	Intended for easy bending up to 90 deg. across the grain.[e] (No bending along the grain.)
No. 3. Quarter-hard .	69 ± 5	8 to 9	54 000 ± 6 000	20 ± 7	For shallow drawing and stamping, where a very smooth surface is required. Bends 180 deg. across the grain. Bends up to 90 deg. along the grain.[e]
No. 4. Soft or planished.............	58 ± 6	9 to 10½	48 000 ± 5 000	30 ± 6	For fairly deep drawing where no sign of surface strain [f] is permissible. Bends 180 deg. both ways of the grain.
No. 5. Dead soft	45 ± 7	10 to 11½	44 000 ± 4 000	39 ± 6	For deep drawing where slight stretcher strains [f] are permissible. Also for drifting. (Erroneously called "extrusion.") Bends 180 deg. both ways of the grain.

[a] Cup depth varies with thickness of strip. For Grade No. 5, dead soft temper, the depth is given approximately by the formula, $D = 10.5$ mm. $+ 6.4$ log "t" (t = thickness in mm.). Other tempers vary in a similar way.

[b] Tensile properties are based on the standard tension test specimen for sheet metals, see Fig. 7 of the Standard Methods of Tension Testing of Metallic Materials (A.S.T.M. Designation: E 8) of the American Society for Testing Materials. 1936 Book of A.S.T.M. Standards, Part I, p. 833.

[c] Elongation in 2 in. varies with thickness of strip. For Grade No. 5, dead soft temper, the percentage elongation $= 41 + 10$ log "t" (t = thickness in mm.). Other tempers vary in a similar way.

[d] For cold rolled strip 0.069 in. and thinner in thickness the Rockwell B hardness range is 96 ± 6, with corresponding increase in tensile strength and drop in depth of cup test.

[e] To bend across the grain means that the crease formed by the bend shall be at right angles to the length of the cold rolled strip. To bend along the grain means that the crease formed by the bend shall be parallel with the length of the cold rolled strip.

[f] See third paragraph of the Appendix as to "age hardening" and the necessity for using steel promptly after rolling where stretcher-strains are to be entirely absent.

From an Appendix to "Standard Specifications for Cold-Rolled Strip Steel — A.S.T.M. Designation: A109-38."

down ingot molds. Channels, beams, angles, plate, etc., rolled from these ingots constitute an important fraction of the total tonnage of steel produced. (Some higher carbon steel shapes and several types of alloy steels are also used for structural purposes in bridges and heavy machinery.)

Boiler plates, rivets, and tubes are usually in the range 0.15 to 0.30% carbon, although higher strength alloy steels are in use in high pressure boilers.
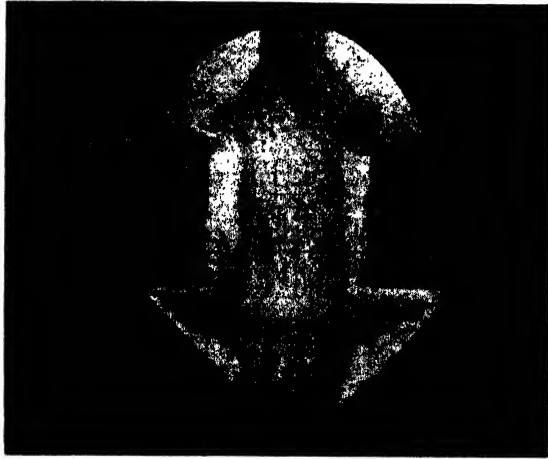
FIG. 102. — DEEPLY ETCHED SECTION OF A RIVETED JOINT TAKEN FROM A CONDEMNED BOILER. (Etched for 5 minutes in a boiling solution of 50% HCl-50% H_2O.) X1.
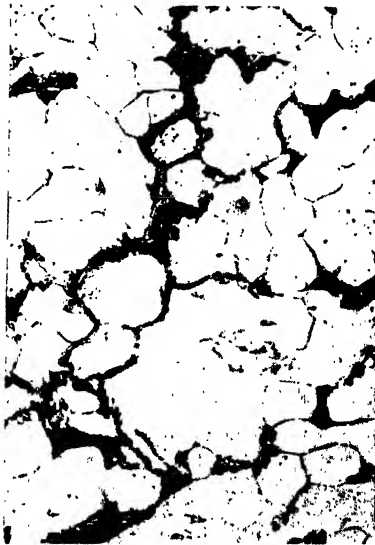
FIG. 103. — MICROSTRUCTURE OF THE RIVET SHOWN IN FIG. 102 TAKEN IN THE REGION OF A CRACK. (Etched with picral.) X200.

Intercrystalline Corrosion Cracking. — One of the recurrent causes of failures in steam boilers is intercrystalline cracking at points of stress concentration in rivets and at rivet holes. A section of a riveted joint taken from a condemned boiler is shown in Fig. 102. Many of the rivet heads of this boiler cracked off when struck with a hammer. Intergranular corrosion is evident in Fig. 103, the microstructure of the rivet taken near a crack.

For many years it was believed that sodium hydroxide in the boiler water was the principal source of trouble and the term " caustic embrittlement " was applied to failures of this type. More recently it has been proved that sodium hydroxide has little effect except when other substances are also present. According to Schroeder, Berk, and O'Brien [23]: ". . . intercrystalline cracking is produced by a solution that tends to form a protective film over the crystal surfaces while it attacks the grain boundaries. This condition can most easily be attained with two dissolved components, but with careful adjustment it may be possible to produce cracking in a solution containing only one dissolved substance." Although sodium silicate is one of the most dangerous additions, titanium oxide and antimony oxide in caustic solutions also promote cracking. Certain non-caustic solutions such as sodium nitrate with additions of manganese compounds also produce failure. Investigation shows that the conditions of temperature and stress have a critical effect on intergranular cracking, which complicates the matter of providing protection when impure boiler waters are used.

The Use of Deep Etching for Macrostructures. — The macrostructure of the riveted joint, Fig. 102, illustrates the use of deep etching methods to reveal " flow lines " and other features of wrought steels. The marked difference in color between the outer layer and the interior of the rivet indicates that it is a rimming steel having a surface layer of relatively pure iron. (See page 178.) The flow lines in the rivet and plates are caused by elongation of inclusions and possibly the microconstituents in the direction of working. This condition is prevalent in all wrought steels and does not, in itself, indicate excessive segregation; however, banding may become quite serious, as illustrated in Fig. 97, resulting in low ductility and toughness of cross fiber, or transverse specimens com-

pared with straight fiber or longitudinal specimens of the same material.

Flow lines are very useful in determining the flow of metal in forging dies and the direction of the fiber in gear teeth, milling cutters, etc. A gear cut from a blank which has been forged or machined from flat stock will have teeth with either straight or cross fiber depending on their location with respect to the direction of rolling. The impact toughness of the teeth having cross fiber will be comparatively low. A gear blank made by upsetting the end of a bar of much smaller diameter will have essentially straight fiber in all teeth, insuring better uniformity of toughness and wear resistance and less warpage in heat treatment. Obviously a gear blank which is machined from bar stock will be more uniform in structure than one from flat stock; however, all teeth will be cross grained.

There are many other useful applications of macroetching in revealing surface defects, quenching and grinding cracks, soft spots, dendritic structure, internal cracks or " flakes," pipe, center porosity, ingot pattern effect, penetration of weld deposits, carburization, and decarburization. (See pages 359, 406, and 422.) Tool steels are very frequently subjected to deep etching tests to detect defects in manufacture and treatment. A series of papers published by The American Society for Metals contains detailed information on the application and interpretation of macroetching.[24]

S.A.E. Carbon Steels. — Another important group of steels containing 0.10 to 0.20% carbon, and essentially the same basic composition as structural steel, is manufactured under melting and ingot making conditions which are conducive to higher quality. These steels are included in the well known Society of Automotive Engineers classification as S.A.E. 1010, 1015, and 1020. They may be considered to be carbon steels of alloy quality and, as such, demand a higher price than ordinary structural steel.

The Society of Automotive Engineers chemical specifications for carbon and free-cutting steels follow in Table XXVI.[25] Although the principal use of the S.A.E. numbers will be made in the chapter on alloy steels, the numbering system will be examined here in order that it may be used in the discussion of carbon steels. The first digit

has the most general significance; for example, the numbers of all of the carbon steels of Table XXVI are of the form 1xxx, the numeral 1 indicating the absence of alloying elements. In Chapter IX, Table XXXVI, it will be found that symbols of the form 2xxx and 5xxx represent nickel and chromium steels respectively. The second digit modifies or extends the meaning of the first; for example, 10xx indicates a plain carbon steel while 11xx indicates a carbon steel with higher sulphur content than usual, and 13xx indicates higher manganese content. The third and fourth digits refer to the carbon content; for example, S.A.E. 1020 contains 0.15 to 0.25% carbon, while S.A.E. 1055 contains 0.50 to 0.60% carbon. The carbon limits are greater than ±0.05% in some cases but purchasers of steel frequently reduce the carbon limits in their specifications to a total range of 0.05%. The letters X and T, which sometimes appear before the numbers, indicate some further modification of the analysis.

Carburizing Steels. — A large proportion of the production of the low carbon S.A.E. steels 1015 and 1020 is used for carburized parts requiring high surface hardness and good general toughness. In the fabrication of such parts hot forging operations, cold forming, or cold heading may be involved. The requirements of the steels include structural uniformity, good hot working properties or cold plasticity, good carburizing characteristics, and ease of hardening after carburizing. Further references to carburizing steels will be found in Chapter X.

Cold Heading Steels. — S.A.E. 1020 is also one of the principal steels used for cold headed bolts, rivets, and allied products. Cold heading has become one of the important mechanical operations

TABLE XXVI.   CHEMICAL COMPOSITIONS OF S.A.E. STEELS

CARBON STEELS

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Max.	Sulphur Max.
1010	0.05–0.15	0.30–0.60	0.045	0.055
1015	0.10–0.20	0.30–0.60	0.045	0.055
X1015	0.10–0.20	0.70–1.00	0.045	0.055
1020	0.15–0.25	0.30–0.60	0.045	0.055
X1020	0.15–0.25	0.70–1.00	0.045	0.055

TABLE XXVI.   (*Continued*)

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Max.	Sulphur Max.
1025	0.20–0.30	0.30–0.60	0.045	0.055
X1025	0.20–0.30	0.70–1.00	0.045	0.055
1030	0.25–0.35	0.60–0.90	0.045	0.055
1035	0.30–0.40	0.60–0.90	0.045	0.055
1040	0.35–0.45	0.60–0.90	0.045	0.055
X1040	0.35–0.45	0.40–0.70	0.045	0.055
1045	0.40–0.50	0.60–0.90	0.045	0.055
X1045	0.40–0.50	0.40–0.70	0.045	0.055
1050	0.45–0.55	0.60–0.90	0.045	0.055
X1050	0.45–0.55	0.40–0.70	0.045	0.055
1055	0.50–0.60	0.60–0.90	0.040	0.055
X1055	0.50–0.60	0.90–1.20	0.040	0.055
1060	0.55–0.70	0.60–0.90	0.040	0.055
1065	0.60–0.75	0.60–0.90	0.040	0.055
X1065	0.60–0.75	0.90–1.20	0.040	0.055
1070	0.65–0.80	0.60–0.90	0.040	0.055
1075	0.70–0.85	0.60–0.90	0.040	0.055
1080	0.75–0.90	0.60–0.90	0.040	0.055
1085	0.80–0.95	0.60–0.90	0.040	0.055
1090	0.85–1.00	0.60–0.90	0.040	0.055
1095	0.90–1.05	0.25–0.50	0.040	0.055

FREE-CUTTING STEELS

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus Range	Sulphur Range
1112	0.08–0.16	0.60–0.90	0.09–0.13	0.10 –0.20
X1112	0.08–0.16	0.60–0.90	0.09–0.13	0.20 –0.30
1115	0.10–0.20	0.70–1.00	0.045 max.	0.075–0.15
1120 *	0.15–0.25	0.60–0.90	0.045 max.	0.075–0.15
X1314	0.10–0.20	1.00–1.30	0.045 max.	0.075–0.15
X1315	0.10–0.20	1.30–1.60	0.045 max.	0.075–0.15
X1330	0.25–0.35	1.35–1.65	0.045 max.	0.075–0.15
X1335	0.30–0.40	1.35–1.65	0.045 max.	0.075–0.15
X1340	0.35–0.45	1.35–1.65	0.045 max.	0.075–0.15

* The Iron and Steel Division contemplates discontinuing S.A.E. 1120 at an early date in view of its gradual displacement in industry by S.A.E. 1115. (Note — Dec. 7, 1937.)

for fabricating steel. In 1933 it was estimated that the capacity of the cold heading branch of the bolt industry was some five billion pieces or 300,000 tons of metal in wire and rod sizes ranging from 0.16 to 0.75 inch.[26]   Compared with cold heading, screw machine production of bolts requires free-machining stock and wastes a large part of the bar, but gives high dimensional accuracy and concentricity. Larger sizes of bolts are commonly upset in hot headers, but the finish does not compare with the cold formed product.

Metals for cold heading must have good plasticity and general soundness including freedom from surface defects. Unfortunately the free-cutting screw stocks are not well suited to cold heading because of their inherent brittleness. The requirements of cold heading steels and developments in cold heading practice are fully discussed by Graham [27] and Smith [28] respectively.

The principal steels used for cold heading are S.A.E. 1010, 1020, 1035, 3135, and 2330. Many nonferrous alloys are also suitable for this purpose. The low carbon steels are suitable for general applications in which high strength is not required or for bolts to be case hardened by carburizing. The medium carbon and alloy steel grades are generally heat treated for high strength.

Free-Cutting Steels. — The S.A.E. low carbon free-cutting bar steels (see Table XXVI) are generally finished by cold drawing. The accurate and uniform gage and the excellent surface finish of drawn bars makes them particularly suited for high speed screw machine stock, and for other machining purposes. The strain-hardening produced by cold finishing also improves the machinability of soft steels which ordinarily tend to flow ahead of the tool, tear, heat excessively, and produce a rough machined surface.

Chips form easily and break short when free-cutting steels are machined. This is due, in part, to the presence of sulphide and other slag particles which tend to harden and embrittle the steel, and to the direct hardening effect of alloying elements such as manganese and phosphorus. Grain size and the condition of the microconstituents also influence machinability.[26] Coarse grained steels are generally easiest to machine but do not, ordinarily, take the best finish. It is found that machinability increases with hardness up to a certain point, after which it drops rather rapidly.

Graham [30] has advanced the following explanation for free-machining qualities. "It might be assumed that the minimum energy required for rupture would be associated with minimum hardness, yet practical experience shows that very soft steels are notoriously difficult to machine. An example of this is S.A.E. 1015 steel. With increased hardness there is a decreased 'mushing' action; that is, the cutting force penetrates a lesser distance into the metal around the tip of the tool. Less volume of metal is deformed in the process of rupture and, although the amount of energy which must be expended per unit volume of metal is increased due to the increased hardness, yet the total energy is diminished by reason of the lesser volume of metal involved. The decrease in volume becomes less with increased hardness until a point is reached beyond which there is no further decrease in the total amount of metal involved in the cutting operation. This is the point of maximum machinability; that is, of minimum total energy required. Beyond this point any increase in hardness is accompanied by a proportionately greater total energy which means a corresponding decrease in free-cutting qualities."

Free-cutting qualities are developed in four different classes of S.A.E. steels, and in a new alloy containing lead. S.A.E. X1015 and X1020 contain 0.70 to 1.0% manganese as the hardening agent. Although these steels are not equal to certain other types in developing free-machining qualities, they are capable of producing carburized parts having good hardenability and excellent core properties. They may be forged if necessary or fabricated by other means more easily than the high sulphur types to be considered.

Each of the three other classes of S.A.E. free-cutting steels contains about 0.10% sulphur or over. Steels of the second class, S.A.E. 1115 and 1120, are similar to X1015 and X1020 except for the higher sulphur content of the former. The MnS and FeS inclusions in these steels effectively embrittle the turnings, preventing the formation of long, tough ribbons which would quickly interfere with the tooling setups of high speed, automatic machine tools.

The carbon steels containing more than 1.0% manganese are given the number 13xx in the S.A.E. classification. S.A.E. X1314 and X1315 are high manganese, high sulphur steels whose machin-

ability is accompanied by excellent hardenability after carburizing. This favorable combination of properties has caused them to be specified more and more to meet the requirements of carburized parts to be machined at high speeds.

The fourth type of S.A.E. free-cutting steel is the familiar Bessemer screw stock, S.A.E. 1112 and X1112, in which high phosphorus and high sulphur are both used, giving even better machinability than that of the high sulphur open hearth grades. According to Epstein and Miller [31]: " Aging steels, since they harden and become embrittled more readily under cold work than non-aging, machine more easily; the chips become brittle under the action of the cutting tool and break off. For this reason Bessemer steel makes a freer cutting screw stock than open hearth steel, the higher N_2 and O_2 in Bessemer steel probably promoting aging."

The ratings by Boston [32] in Table XXVII indicate the superiority of the Bessemer screw stock.

It will be recalled that small percentages of lead are used for imparting free-machining qualities to brass and other nonferrous alloys. Only recently lead has been added to steel for this purpose. It has been found that additions of 0.15 to 0.25% lead to low carbon free-cutting steels effect a further improvement in machinability. Ordinary metallographic preparation does not reveal the presence of lead in the microstructure, and chemical analyses of ingot sections indicate that it is very well distributed. According to Nead, Sims, and Harder [33], who have recently reported the properties of lead-bearing steels: ". . . it appears probable that the lead is present in a submicroscopic state of division, analogous to the dissemination of particles in a precipitation-hardened alloy or a grain size controlled steel, i.e., in a very different order of size and far more uniform dissemination than is the case in leaded brass."

Unlike the sulphur and phosphorus additions to screw stock, lead does not reduce the ductility and toughness of steel; in fact, the influence of lead on all mechanical properties has been shown to be slight and for this reason it is very likely that it will be used to improve the machinability of the higher carbon and alloy steels.

Bessemer Steels. — The relative costs of open hearth and Bessemer steels depend somewhat upon the price of steel scrap, one of

TABLE XXVII. MACHINABILITY RATINGS OF COLD DRAWN
BARS FOR SCREW MACHINE USE **

S.A.E. No.	Rating, %	S.A.E. No.	Rating, %
Screw Machine Stock		*Heat Treating Grades*	
1120 (leaded)	140	1035	55–60
X1112	140	1040	55–60
1112	100	1045 *	55–60
X1314	90–95	4130 *	40–50
X1315	90–95	4140 *	40–50
1120	70–75	3140 *	40–50
X1335	70–75	6135 *	40–50
X1330	70–75	6140 *	40–50
Carburizing Steels		6150 *	40–50
X1020	50–65	1335 *	40–50
1025	50–65	1350 *	30–40
1020	50–65	2330 *	30–40
1015	50–65	2345 *	30–40
6120 *	45–50	3140	30–40
6115	45–50	2345	30–40
3115	45–50	52100 *	30–40
4615 *	45–50		
2315 *	45–50		
3215	45–50		

* Annealed.

** " These values are for high speed steel tools turning, forming, drilling, threading, and cutting off when a conventional cutting fluid is used. The ratings are based on the machinability of the Bessemer screw stock, S.A.E. 1112, as 100%. This corresponds to a cutting speed of about 150 feet per minute. Annealed means the bar was annealed before cold drawing. If not so indicated they are cold drawn directly from the hot rolled bars.

" The values given are to serve only as a guide and should be changed to suit specific operations and conditions. These values are based on the results of an extensive survey."

Courtesy of O. W. Boston, Metals Handbook.[32]

the principal raw materials of the open hearth process, compared with the price of pig iron, from which Bessemer steels are made. Bessemer steels are, at times, more expensive than ordinary open hearth steels. Because of the inability of the acid converters to remove phosphorus or sulphur from the pig iron, and because of the retention of relatively large amounts of FeO as slag particles, Bessemer steels have certain inherent characteristics not found in open hearth steels.[54] The exceptional machinability of Bessemer

steels has already been discussed.   Their good forge welding char-
acteristics facilitate the production of low carbon steel pipe.   Other
Bessemer products include low carbon sheet, wire, and structural
steel, and higher carbon reinforcing bars and structural steel with
up to 0.35% carbon.   Despite the stiffening and embrittling effects
of the oxides, sulphides, and possibly nitrides present, Bessemer
steels withstand a moderate amount of cold work, and in this con-
dition they develop surprisingly good mechanical properties.

The inherent advantages of Bessemer over open hearth steel for
certain purposes are being recognized more and more, so that the
trend of replacement of Bessemer plants by open hearths has
stopped, and the Bessemer steels are in a position to compete with
several open hearth grades.   The available capacity for the pro-
duction of Bessemer steel in 1938 was approximately one-tenth
that of open hearth steel, but six times that of electric steel.

Cold Drawing. — Brief mention has been made of cold drawing
of many grades of carburizing and free-cutting bar steels.   Further
consideration of the cold finishing processes and the properties
developed is necessary to appreciate the important position of these
steels in industry.   Rounds, flats, hexagons, and many special shapes
up to approximately three inches in cross section may be cold drawn
from oversize hot rolled bars.   Special sections drawn to finished
dimensions effect great savings in machining costs.   The drawing
process is better adapted than rolling to the production of these
shapes because of the relatively low cost of drawing dies compared
with rolls.

In addition to cold drawing through dies, cold finishing may be
accomplished by turning and polishing hot rolled bars, or where
exceptionally good surface and concentricity is required either hot
rolled or cold drawn bars may be ground to size.   Thus it is possible
to obtain any desired accuracy of dimensions and perfection of fin-
ish, with mechanical properties corresponding to either the hot
rolled or the cold drawn state.

The most widely used grades for cold drawing are S.A.E. steels
1112, X1112, 1020, X1020, 1035, 1040, 1045, 1120, X1314,
X1315, and X1335.[35]   Higher carbon and alloy grades are also
drawn for special purposes.

Armour's data [36] in Table XXVIII show the effect of cold draw-
ing on the properties of a low carbon and an alloy steel. Although
the reductions in the dies are small, seldom over 15%, the effect
on the mechanical properties is considerable. Soft, low carbon

TABLE XXVIII.   EFFECT OF COLD DRAWING ON MECHANICAL
PROPERTIES

0.16% CARBON OPEN HEARTH STEEL

	Yield Point lb./sq. in.	Ultimate Strength lb./sq. in.	Elongation in 2 in., %	Red. of Area, %	Brinell Hardness
Hot rolled bar, 1 3/32 in.	44,500	69,250	37	65	137
Reduced 5% to 1 1/16 in.	61,050	81,750	19.5	58	170
Change	+35%	+18%	−47%	−10%	+24%
Reduced 10% to 1 1/32 in.	84,500	88,250	16.5	56	179
Change	+90%	+27%	−55%	−13%	+30%
Reduced 15% to 1 in.	90,100	94,250	15.5	55	187
Change	+102%	+36%	−58%	−15%	+36%

0.35% CARBON, 1.25% NICKEL, 0.60% CHROMIUM STEEL

Heat treated bar, 1 in.	89,150	107,900	26	67	228
Hot rolled bar, 1 1/16 in.	102,500	131,550	15.5	41	248
Same, reduced 10% to 1 in.	116,150	146,050	9.5	35	262
Change	+13%	+11%	−32%	−14%	+5%

Courtesy of J. D. Armour, Metal Progress. [36]

steels, especially Bessemer steels, are most subject to strengthening
by cold work. As demonstrated for sheet steels, cold work elimi-
nates the yield point elongation which is characteristic of the hot
rolled or soft annealed metal. The tests reported in Table XXVIII
were performed on 0.505 inch diameter tensile specimens which
were machined from the larger bars. The marked changes in prop-
erties indicate that cold drawing is not merely a surface effect.

The high yield point which was reported for the bar reduced 15%

is probably greater than could be obtained by any heat treatment applied to this low carbon steel. However, a yield strength of 90,-000 lb. per sq. in. can easily be obtained by heat treatment of a higher carbon or alloy steel; for example, a nickel-chromium steel of the type used by Armour may be quenched in oil from 1500° F. and reheated to 1200° F. to develop approximately the same yield strength. It would have a lower ratio of yield strength to tensile strength, greater hardness, and better ductility than the cold drawn carbon steel; however, the machinability of the cold drawn carbon steel bar is superior, and its properties are satisfactory for many purposes where the increased cost of heat treated alloy steels is not warranted.

Cold drawing of seamless steel tubing, like cold drawing of rod, is a process which grew out of the much older art of wire drawing.[37] Only the best grades of welded tubing compare with drawn tubing for strength, uniformity, and reliability.

Carbon Steels for Water Quenching. — Aside from the important group of low carbon carburizing steels mentioned above, the products discussed so far are not subject to appreciable strengthening or hardening by heat treatment. The lower limit for susceptibility to heat treatment may be arbitrarily set at 0.20% carbon. This does not mean that homogenization and strengthening of the low carbon steels is impossible, but that common commercial practice does not include heat treatment for strengthening.

The S.A.E. carbon steels which may be hardened by quenching in water or oil are those in the second to fifth groups in Table XXVI, page 188. Details of the heat treatment of these steels will be deferred to the next chapter.

S.A.E. 1030 is typical of the water quenching carbon steels which are used mainly for forgings, seamless tubing, heavy stamped or pressed parts, or machined parts requiring moderate strength or wear resistance. These steels are readily forged, machined, or cold formed as in bolt making, and after heat treatment they have notably better mechanical properties than low carbon steels. It will be shown later that steels in this general carbon range are usually shallow hardening; therefore, heat treated articles should be restricted to those of small section if uniform strength and hardness are desired. Higher manganese contents, as in S.A.E. X1025, improve

the hardenability. It is possible to apply a hard surface case to 0.25 to 0.35% carbon steels by carburizing or cyaniding for applications in which surface hardness must be accompanied by moderately high core strength; however, the lower carbon steels are ordinarily sufficiently strong for carburized parts and have the advantage of greater toughness and capacity for cold deformation compared with the higher carbon varieties.

Medium Carbon Steels for Water or Oil Quenching. — The S.A.E. steels from 1035 to 1055 inclusive in Table XXVI may be hardened in water or oil depending upon the size and intricacy of the part. As the carbon content increases and as the size of the part decreases or its cross section becomes complicated by drastic dimensional changes, sharp corners, screw threads, key ways, etc., the tendency is to use oil to reduce the danger of quenching cracks.

S.A.E. 1040 is typical of the stronger medium carbon steels which may be hardened in water unless the characteristics of the section require oil. The automotive industry has made wide application of this steel for crankshafts, connecting rods, tubing, axles, brake levers, bolts, etc.

Gears and heavier carbon steel forgings are made from higher carbon steels such as S.A.E. 1050, especially when rather high strength is required throughout the entire section. Steels in this carbon range are essentially oil hardening, except in uniformly heavy sections where water may be used safely and to better advantage.

Medium Carbon to Eutectoidal Steels. — S.A.E. 1055, the lowest carbon steel ordinarily used for springs, has been applied to both hard drawn and heat treated coil springs. The steels in groups four and five of Table XXVI may be classified as spring steels and lower carbon tool steels. They are commonly used as tools for wood working, and for discs, harrows, plowshares, shovels, rake teeth, mower knives, etc., in the agricultural industry.[38]

Steels similar to those in the fourth group constitute the important tonnage steels used for rails and wheels by the railroads.[39] The most usual carbon range for these products is 0.65 to 0.75% and the manganese contents are approximately the same as the carbon.

Eutectoidal Steels for Springs and Wire. — S.A.E. 1090 and 1095 are used extensively for both leaf and coil springs, as well as

for tools of various kinds including punches and dies, chisels, saws, hammers, etc.   Other tool steels have carbon contents up to about 1.5% ; however, further classification of these steels will be deferred to Chapter XV.

High strength wire for springs, wire rope, piano wire, and structural purposes is drawn from high grade open hearth or electric
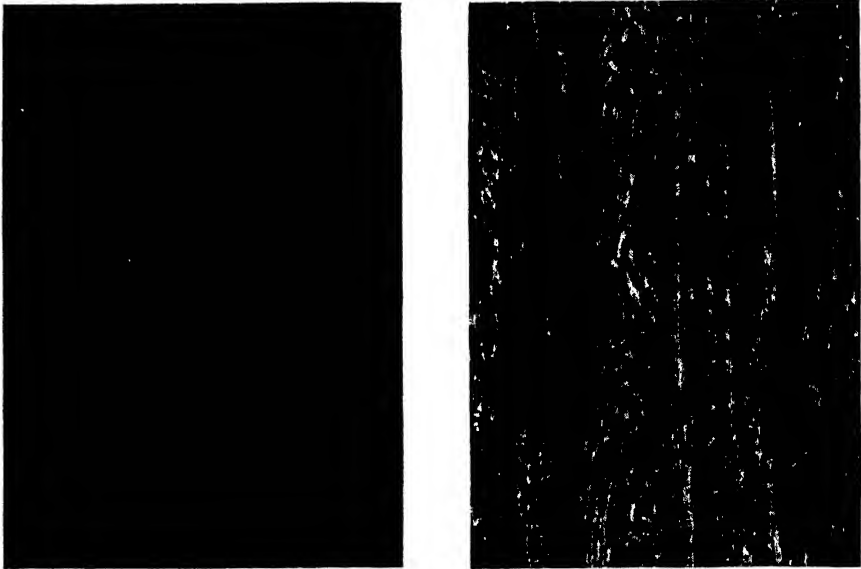
Left.   Fig. 104. — Longitudinal Section of a Hard Drawn Steel Wire Containing 0.74% C, 1.13% Mn, 0.16% Si. (Etched with 3% nital.)   X100.
Right.   Fig. 105. — Same as Fig. 104.   X1500.

steel containing 0.40 to 0.95% carbon.   The eutectoidal compositions containing 0.75 to 0.85% carbon are most often used.   The high strength of these wires is the result of cold work rather than heat treatment.   Although equivalent strength can be obtained by heat treating wire of the same composition, the first major application of heat treated wire for suspension bridge cables resulted in failure [40], whereas cold drawn wires have been highly successful for bridge wire, wire ropes, and similar applications.   The heat treated wire used in these cables had a comparatively rough surface with occasional surface cracks, whereas the surface of the cold drawn wire was smooth and highly compacted.   The galvanized coating which

was used on both types penetrated the fissures in the heat treated wire. This condition introduced regions of stress concentration which contributed to its failure.

The microstructures of cold drawn eutectoid steels consist of very fine and highly deformed pearlite grains, as shown in Figs. 104 and 105. The " patenting " heat treatment (see page 203), which is applied before the wire drawing operations, gives a very fine, equiaxed pearlitic structure which is sufficiently plastic to withstand cold reduction but capable of developing high strength after drawing. Only at high magnification can laminations be observed in the pearlite, and then only in certain grains. The term " sorbitic pearlite " has often been applied to structures in which the pearlite is very finely lamellar and appears to contain cementite in the form of small particles rather than plates.

Patented wire drawn to 240,000 lb. per sq. in. tensile strength has a ductility of about 3 % elongation in 10 inches.[41] The ductility may be increased and the tensile strength decreased slightly by the stress relieving effect of galvanizing at about 850° F. Thum [41] describes the processing of wire for bridge cables.

Wire for flexible cables or wire ropes varies in strength from about 150,000 to 270,000 lb. per sq. in. depending on the carbon content and amount of cold reduction. Tensile strengths over 300,-000 lb. per sq. in. are obtained in very small diameter piano wires.

REFERENCES

1. Cleaves, H. E., and Thompson, J. G., " The Metal — Iron," The Engineering Foundation, McGraw-Hill Book Company, New York (1935)
2. Green, A. W. F., " Wrought Iron," Metals Handbook, 457, Am. Soc. for Metals, Cleveland (1939)
3. Epstein, S., " The Alloys of Iron and Carbon, Vol. I — Constitution," The Engineering Foundation, McGraw-Hill Book Company, New York (1936)
4. Mehl, R. F., and Wells, C., " Constitution of High-purity Iron-carbon Alloys," Trans. Am. Inst. Mining and Met. Eng., 125, 429 (1937)
5. Sauveur, A., " The Metallography and Heat Treatment of Iron and Steel," McGraw-Hill Book Company, New York (1935)
6. Epstein, S., Nead, J. H., and Halley, J. W., " Choosing a Composition for Low-alloy High-strength Steel," Trans. Am. Inst. Mining and Met. Eng., 120, 309 (1936)
7. Stoughton, B., " The Metallurgy of Iron and Steel," 391, McGraw-Hill Book Company, New York (1934)
8. Walters, F. M. Jr., and Wells, C., " Alloys of Iron, Manganese and Carbon, Part XV," Trans. Am. Soc. for Metals, 24, 359 (1935)

9. Herty, C. H. Jr., " The Physical Chemistry of Steel Making," Mining and Metallurgical Advisory Boards, Pittsburgh (1934)

10. " Distribution of Finished Steel in 1937," Steel, **102,** 23 (March 14, 1938)

11. Griffis, R. O., Kenyon, R. L., and Burns, R. S., " The Aging of Mild Steel Sheets," Year Book Am. Iron and Steel Inst. (1933)

12. Davenport, E. S., and Bain, E. C., " The Aging of Steel," Trans. Am. Soc. for Metals, **23,** 1047 (1935)

13. Hayes, A., and Burns, R. S., " The Cold Rolling of Mild Steel Sheets and Strip," Trans. Am. Soc. for Metals, **25,** 129 (1937)

14. Olt, T. F., " Production and Quality Control of Sheets for Automobile-Body Fabrication," Trans. Am. Soc. for Mech. Eng., **59,** 185 (1937)

15. Fleming, W. R., " On the Manufacture of Rimming Steel," Trans. Am. Soc. for Metals, **22,** 532 (1934)

16. Halley, J. W., and Washburn, T. S., " Distribution of Metalloids in Rimmed-steel Ingots," Trans. Am. Inst. Mining and Met. Eng., **131,** 195 (1938) — and — McCutcheon, K. C., and Chipman, J., " Evolution of Gases from Rimming-steel Ingots," 206

17. Sauveur, A., " Rimmed vs. Killed Steel," Metals and Alloys, **9,** 343 (1938)

18. Kenyon, R. L., " The Importance of Grain Size of Sheet Steel for Deep Drawing," Trans. Am. Soc. for Metals, **22,** 1099 (1934)

19. " Standard Definitions and Terms Relating to Methods of Testing — A.S.T.M. Designation E6–36," A.S.T.M. Standards, Part iI, 1429 (1936)

20. Kenyon, R. L., and Burns, R. S., " Autographic Stress-Strain Curves of Deep-Drawing Sheets," Trans. Am. Soc. for Steel Treating, **21,** 577 (1933)

21. Hayes, A., and Griffis, R. O., " Non-aging Iron and Steel for Deep-Drawing," Metals and Alloys, **5,** 110 (1934)

22. Daniloff, B. N., Mehl, R. F., and Herty, C. H. Jr., " The Influence of Deoxidation on the Aging of Mild Steels," Trans. Am. Soc. for Metals, **24,** 595 (1936)

23. Schroeder, W. C., Berk, A. A., and O'Brien, R. A., " Intercrystalline Cracking of Steel in Aqueous Solutions," Metals and Alloys, **8,** 320 (1937)

24. Trans. Am. Soc. for Steel Treating, **21,** 289, 310, 343, and 354 (1933)

25. S.A.E. Handbook, Soc. of Automotive Eng. (1938)

26. Pulsifer, H. B., " Materials for Cold Heading," Metal Progress, **23,** 13 (March 1933)

27. Graham, H. W., " Requirements of Steel for Cold Heading," Metal Progress, **24,** 22 (Aug. 1933)

28. Smith, R. H., " Cold Heading — Bolts, Rivets, Nails," Trans. Am. Soc. for Metals, **25,** 84 (1937)

29. " Some Factors Affecting Machinability of Steels and Other Alloys," Metal Progress, **29,** 31 (Jan. 1936)

30. Graham, H. W., " Grain-Size in Relation to Machinability and Other Properties of Bessemer Screw Steel," Trans. Am. Soc. for Metals, **22,** 926 (1934)

31. Epstein, S., and Miller, H. L., " Aging in Iron and Steel," Metals Handbook, 602, Am. Soc. for Metals, Cleveland (1939)

32. Boston, O. W., " Machinability of Steel," Metals Handbook, 891, Am. Soc. for Metals, Cleveland (1939)

33. Nead, J. H., Sims, C. E., and Harder, O. E., " Properties of Some Free-Machining, Lead-Bearing Steels," Metals and Alloys, **10,** 68 (1939)

34. Henning, C. C., " Manufacture and Properties of Bessemer Steel," Trans. Am. Inst. Mining and Met. Eng., **116,** 137 (1935)

35. Beck, J. E., " Cold Forming Processes — Drawing Rods and Bars," Trans. Am. Soc. for Metals, **25,** 98 (1937)

36. Armour, J. D., " Cold Finished Bars, Properties and Applications," Metal Progress, **27,** 43 (May 1935)

37. Knerr, H. C., " Cold Drawing Seamless Steel Tubes," Trans. Am. Soc. for Metals, **25,** 114 (1937)

38. Bornstein, H., " The Wear of Farm Tools," Metals and Alloys, **9,** 291 (1938)

39. Gillett, H. W., " Metallurgy of Rail Steels — Developments in Last 25 Years," Metals and Alloys, **7,** 243 (1936)

40. Swanger, W. H., and Wohlgemuth, G. F., " Failure of Heat-Treated Steel Wire in Cables of the Mt. Hope, R. I., Suspension Bridge," Proc. Am. Soc. for Testing Mats., **36,** Part II, 21 (1936)

41. Thum, E. E., " Manufacture of High Strength Wire," Metal Progress, **22,** 27 (July 1932)

HEAT TREATMENT OF STEEL

The art of hardening steel by forge fire methods has been practiced for many centuries. Modern contributions to the art include adequate heating media, accurate control of the heating and cooling cycle, and sensitive methods of testing the mechanical properties of the product. The requirements of the automotive industry for quantity production of properly heat treated parts have been largely responsible for the application of scientific methods to heat treatment.

In this chapter the terminology and elementary principles of heat treatment of steel will be followed by a more detailed study of the hardening process and the factors which influence hardenability, including the effect of chemical composition and grain size. Heat treating operations and equipment will then be discussed briefly from the standpoint of heating media, quenching baths, and the special processes for local heating and quenching.

Definitions of Heat Treating Processes as Applied to Steels. — The following definitions, condensed from " Metals Handbook "[1] will be used in this text. The statements in brackets are the author's.

Full Annealing — Heating iron-base alloys above the critical temperature range, holding above that range for a proper period of time, followed by slow cooling to below that range. [This definition does not apply to austenitic stainless steels and other alloys which have no critical range.]

Note — the annealing temperature is generally about 100° F. above the upper limit of the critical temperature range, and the time of holding is usually not less than 1 hour for each inch of section of the heaviest objects being treated (see tool steel section in this Handbook for tool steel practice). The objects being treated are ordinarily allowed to cool slowly in the furnace. They may, however, be removed from the furnace and cooled in some medium which will prolong the time of cooling as compared to unrestricted cooling in the air.

Process Annealing — Heating iron-base alloys to a temperature below or close to the lower limit of the critical temperature range followed by

cooling as desired. [This heat treatment is commonly applied in the sheet and wire industries and the temperatures generally used are from 950°–1300° F.]

Normalizing — Heating iron-base alloys to approximately 100° F. above the critical temperature range followed by cooling to below that range in still air at ordinary temperature. [Hypereutectoid steels may be normalized above Acm but are annealed above Ac_1.]

Patenting — Heating iron-base alloys above the critical temperature range followed by cooling below that range in air, or in molten lead or a molten mixture of nitrates or nitrites maintained at a temperature usually between 800° and 1050° F., depending on the carbon content of the steel and the properties required of the finished product.

Note — This treatment is applied in the wire industry to medium or high carbon steel as a treatment to precede further wire drawing. [See page 199.]

Hardening — Heating and quenching certain iron-base alloys from a temperature either within or above the critical temperature range for the purpose of producing a hardness superior to that obtained when the alloy is not quenched. Usually restricted to the formation of martensite. [Hypoeutectoid steels are nearly always hardened from above the critical temperature range while hypereutectoid steels may be hardened from above Ac_1 but below Acm.]

Tempering — (also termed drawing) — Reheating hardened steel to some temperature below the lower critical temperature, followed by any desired rate of cooling. [See page 211.]

Note — Although the terms "tempering" and "drawing" are practically synonymous as used in commercial practice, the term "tempering" is preferred.

Spheroidizing — Any process of heating and cooling steel that produces a rounded or globular form of carbide.

Note — The spheroidizing methods generally used are:

(1) Prolonged heating at a temperature just below the lower critical temperature, usually followed by relatively slow cooling. [See page 213.]

(2) In the case of small objects of high carbon steels, the spheroidizing result is achieved more rapidly by prolonged heating to temperatures alternately within and slightly below the critical temperature range.

(3) Tool steel is generally spheroidized by heating to a temperature of 1380°–1480° F. for carbon steels and higher for many alloy tool steels, holding at heat from 1–4 hours and cooling slowly in the furnace.

Hardening of Steel. — In addition to the above mentioned contributions to the art of steel treating, modern physical metallurgy has made notable progress in the development of the theoretical aspects

of the hardening of steel. It is now possible to separate and explain many factors which were formerly recognized only as mysterious and undefinable properties of the steel or of the heating or quenching media.

To completely harden ordinary carbon steel it is only necessary to heat it above its critical temperature, Ac_3, and cool it sufficiently fast to prevent the formation of ferrite or pearlite. Although the resulting structure, known as martensite, is not unduly complex, the reason for its high hardness is still a controversial matter.

Whenever the heat treatment of steel is under discussion the subject of grain size inevitably receives considerable attention. In order to simplify the study of heat treatment, the basic principles will first be considered independent of the effects of grain size, to be followed later by a detailed examination of the supplementary effects associated with grain size.

Martensite. — The normal microstructure of a slowly cooled or annealed low carbon steel was shown in Figs. 88 and 89, page 166. The manner in which the ferrite and pearlite formed on cooling was described (using a 0.20% carbon steel) on page 163 and should be reviewed at this time.

A ⅞ inch diameter sample of a 0.15% carbon steel was heated to 1660° F., held for 30 minutes, and quenched in cold water, giving the needlelike martensitic structure illustrated in Fig. 106. The marked change in appearance, compared with Fig. 89, and the increase in Brinell hardness from 120 to 201 indicate that rapid cooling has greatly modified this steel. It will be found that the properties of steels with higher carbon contents are even more drastically affected by cooling rate from above the critical temperature range.

The carbon which was present as Fe_3C in the pearlite of Figs. 88 and 89 was dissolved in the austenite grain structure upon heating to 1660° F. No doubt rapid cooling in cold water tended to preserve the structure existing at the quenching temperature; however, it is known that the austenite in hypoeutectoid carbon steels changes from the face-centered cubic to the body-centered cubic structure so rapidly upon cooling that even water quenching is unable to prevent it. Because of the greatly reduced solubility of carbon in B.C.C. compared with F.C.C. iron, it might be expected that carbon as

Fe$_3$C would be precipitated during this transformation, a supposition which is favored by some authorities. According to a more widely accepted viewpoint, based on extensive experimental evidence, martensite is a supersaturated solid solution of carbon in alpha (B.C.C.) iron.*

It should be noted that the carbon content of the martensite will vary with the carbon content of the steel and will be equal to the carbon content of eutectoid and hypoeutectoid steels providing the quench has been successful in converting all the metal to martensite. Hardened hypereutectoid steels generally consist of excess iron carbide or cemented particles associated with a martensitic matrix whose carbon content is about 0.80%.

There is even less agreement on the fundamental cause for the hardness of martensite than there is on its structural constitution. Sauveur [3] in discussing the reasons for the hardness of martensite mentions the following possibilities:
" (1) — Because it is a solid solution. (2) — Because it is a supersaturated solid solution. (3) — Because of the fineness of its grains. (4) — Because of the distortion of the crystal lattices. (5) — Because of the presence of minute particles of the carbide. (6) — Because of internal strain."

FIG. 106. — 0.15% CARBON STEEL. QUENCHED IN COLD WATER FROM 1660° F. (Etched with picral.) Brinell 201. X500.

The first hypothesis fails to explain the unique hardness of martensite, as compared with austenite of similar carbon content. The third hypothesis is not supported by experimental evidence. The four remaining factors are not contradictory, but complementary,

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* X-ray evidence shows that a transitional body-centered tetragonal structure also exists in newly quenched martensitic structures.[2] Retained austenite may also be present in the higher carbon steels after quenching.

and are in accord with the primary concept of the slip interference theory, that increased disorganization in any form leads to increased strength and hardness.

X-ray evidence indicates that, in newly quenched martensite, iron and carbon atoms are competing for positions in the lattice, forcing
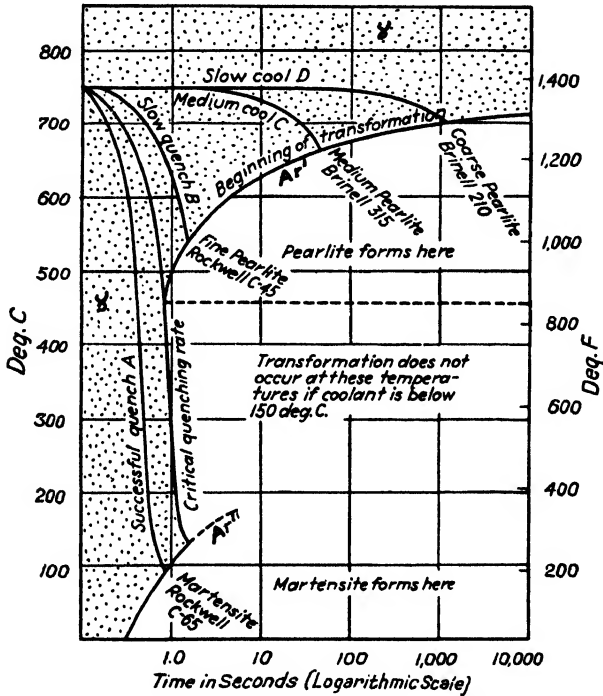
FIG. 107. — CHART SHOWING SCHEMATICALLY THE TIME AND TEMPERATURE OF INITIAL TRANSFORMATION OF A EUTECTOID STEEL AS COOLED AT VARIOUS UNIFORM RATES. (*Courtesy of E. C. Bain, Metals and Alloys.*[10] The shading of the austenitic region has been added by the author.)

changes in the cell dimensions. Such a condition is one of *internal stress*, is a condition of *supersaturation*, and involves a *lattice distortion*. The ejection of iron and carbon atoms, in the form of carbide, is imminent. In time, carbide particles may form, relieving somewhat the above conditions, but introducing hardening through keying, i.e., through lattice distortion localized along slip planes.

Structures Obtained upon Cooling a Eutectoid Steel at Various Rates. — The products formed by cooling carbon steels very rapidly or very slowly are not unduly difficult to identify or comprehend. Intermediate cooling rates are frequently encountered in commercial heat treating practice and the resulting structures are often quite complex.

Referring to Fig. 87, page 165, it will be recalled that the actual Ar_1 transformation temperature was slightly depressed below the equilibrium transformation temperature when a steel was cooled at the rate of $\frac{1}{2}$ ° C. per minute. Considering a eutectoid steel, the depression below the equilibrium transformation temperature increased with the cooling rate as follows: 8° at $\frac{1}{8}$ ° C. per minute, 11° at $\frac{1}{2}$ ° C. per minute, and 16° at 2° C. per minute.[4] It is experimentally true that higher cooling rates cause even greater depression of the transformation temperature, and, as shown in Fig. 107,* the transformation to pearlite occurs in a shorter period of time. (See temperature-time cooling curves D, C, and B.) The time scale for Fig. 107 is logarithmic in order to show detail at both high and low cooling rates on the same chart. This accounts for the curvature in the uniform rate cooling curves.

The microstructures obtained by cooling small samples of a 0.72% carbon steel at various rates will be useful in interpreting Fig. 107. The slowly cooled or annealed structure of Fig. 108 is similar to Fig. 94 except for the presence of a very small amount of pro-eutectoid ferrite at the grain boundaries. (Other fields in this specimen were free from ferrite.) The cooling rate approximates Bain's rate D and the resulting hardness, Brinell 217, is close to that given in Fig. 107. Cooling in air resulted in the slightly harder microstructure shown in Fig. 109. It is apparent that the cementite is no longer present as coarse plates but in a more finely divided form. The cooling rate was between D and C.

In order to obtain intermediate cooling rates between quenching and air cooling (A and C), a sample was partly submerged in water.

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* Figs. 107 and 123 are from the work of E. C. Bain and associates whose researches on the hardening of steel have been published in a series of papers beginning in 1930.[5, 6, 7, 8, 9, 10] The experimental results and the method of presentation developed in these investigations will be used freely in the following sections.

FIG. 108   FIG. 109   FIG. 110

FIG. 111   FIG. 112   FIG. 113

FIGS. 108 TO 119 SHOW THE STRUCTURES OBTAINED ON HEAT TREATING A STEEL
WROUGHT STEEL PASSENGER CAR WHEEL). ALL SAMPLES WERE COOLED FROM
PICRAL. ALL MAGNIFICATIONS X500. (*Courtesy of R. M. Downes, former student at*
  FIG. 108. — ANNEALED. Rockwell "C" 18. Brinell 217.
  FIG. 109. — NORMALIZED (AIR COOLED IN $\frac{1}{2}$ INCH SECTION). Rockwell "C" 22.
  FIG. 110. — QUENCHED BY PARTIAL SUBMERSION IN WATER. Rockwell "C" 41.
  FIG. 111. — QUENCHED IN OIL. Rockwell "C" 63.
  FIG. 112. — QUENCHED IN WATER. Rockwell "C" 66.
  FIG. 113. — REHEATED TO 600° F. (FIGS. 113 TO 118 TAKEN FROM FULLY HARD-
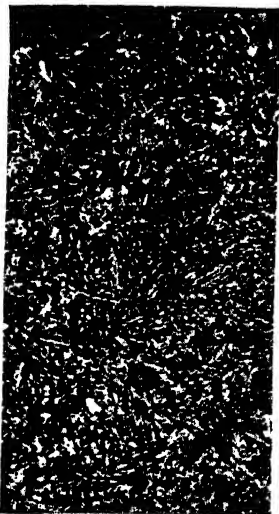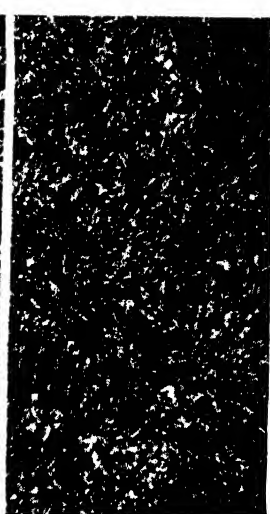
FIG. 114   FIG. 115   FIG. 116
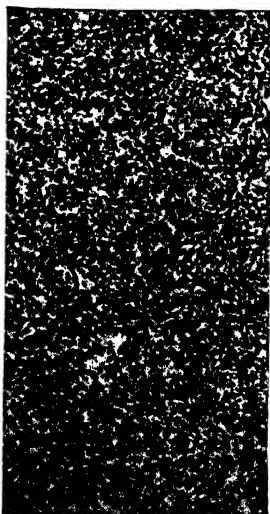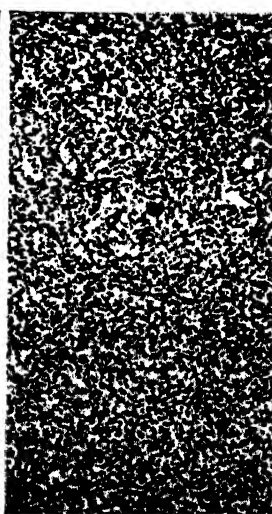
FIG. 117   FIG. 118   FIG. 119

CONTAINING 0.72% CARBON AND 0.75% MANGANESE (TAKEN FROM THE TREAD OF A 1400° F. IN THEIR INITIAL HEAT TREATMENT. ALL SAMPLES WERE ETCHED WITH *Purdue University*.)

> FIG. 114. — REHEATED TO 800° F.   Rockwell " C " 45.
> FIG. 115. — REHEATED TO 900° F.   Rockwell " C " 39.
> FIG. 116. — REHEATED TO 1000° F.   Rockwell " C " 33.
> FIG. 117. — REHEATED TO 1200° F.   Rockwell " C " 22.
> FIG. 118. — REHEATED TO 1300° F.   Rockwell " B " 98.
> FIG. 119. — REHEATED TO 1300° F. FOR 16 HOURS.   Rockwell " B " 86.

At a certain position above the water line the hardness was found to be Rockwell " C " 41 and the microstructure, Fig. 110, was still predominately pearlitic. We may assume that the cooling rate was slightly lower than rate B. The fine pearlite produced by this means is not resolvable at X500 but very high magnification shows that it is at least partially lamellar. The white constituent in Fig. 110 is martensite. Another sample which was quenched in oil was hardened almost to its limit. Its microstructure was predominately martensitic, Fig. 111, although black areas of fine pearlite are also shown. The cooling rate for this sample was slightly less than the critical cooling rate indicated in Fig. 107. The fully martensitic structure of the water quenched sample, Fig. 112, is quite similar in appearance to Fig. 106 and its cooling rate approximates rate A. According to Fig. 107 this sample remained austenitic until a low temperature was attained, whereupon it transformed to martensite at the Ar'' critical temperature. It is for this reason that steel should be kept in the quenching solution until it is well below the hardening temperature range which may be 100° to 400° F.

If it were possible to cool even faster than rate A, one should be able, according to Fig. 107, to retain austenite down to room temperature. It is impossible to cool carbon steels sufficiently fast by commercial quenching treatments to prevent transformation of a considerable portion of the austenite to martensite; however, when the critical cooling rate is sufficiently low the proportion of retained austenite may be sufficient to reduce the hardness well below the maximum attainable value. When this happens it is an indication that less drastic quenching will give full hardness. A more detailed study of the transformation of austenite on heat treatment will follow on page 218.

It should now be apparent that the duplex structures shown in Figs. 110 and 111 are the result of transformation at two different temperature ranges, Ar' and Ar''. In other words, as the critical cooling rate is approached, a part of the austenitic structure may change to fine pearlite at about 1000° F. while the balance of the austenite remains unchanged down to perhaps 300° F., whereupon it transforms to martensite. Split transformation structures occur quite frequently in hardened tools and machine parts. They may

or may not be detrimental depending on whether maximum hardness and strength are required, or whether hardness may be sacrificed somewhat for toughness. Ordinarily only the more slowly cooled interior portions will contain appreciable amounts of the relatively soft fine pearlite, consequently wear resistance at the surface need not be deficient even though the part does not harden through to the center.

Definition of Critical Cooling Rate. — The critical cooling rate may be defined as the *lowest rate* * which will give a fully hardened martensitic structure essentially free from fine pearlite. (From 1 to 3 % fine pearlite is often allowed.) It will be shown later that the three most important factors which influence this characteristic are carbon content, alloy content, and grain size at the quenching temperature. It is important to note further that *critical* cooling rate is a property of the steel in that it does not depend upon the quenching medium, dimensions of the piece, or other factors which may affect the *actual* cooling rate during a given quench. It will, however, vary with grain size, solubility of carbides, and other factors associated with the nature and condition of the steel.

Tempering or Drawing after Hardening. — Hardened steels are seldom used without tempering, either at low temperatures such as 400° F. at which the quenching stresses are relieved without lowering the hardness appreciably, or at higher temperatures up to about 1200° F. at which various desirable combinations of hardness, strength, ductility, and toughness are obtained.

Fully hardened samples of the 0.72% carbon steel described above were tempered at various temperatures, giving the microstructures shown in Figs. 113 to 119. The general terms " sorbite " or " tempered martensite " will be applied to the products of decomposition of martensite. Whether we assume that the carbon is present in solid solution in the martensite or as highly dispersed carbide particles makes little difference, for the carbide particles which form, or are already present, grow to progressively larger sizes as the tempering temperature is increased. The resulting decrease in hardness is caused by the greatly reduced *number of particles* within the grains and not by any change in *grain size*. Although

* Generally measured at a temperature of 1000° F. or slightly higher.

grain boundaries are difficult to show in the highly dispersed and fast etching tempered structures, it is generally believed that the grain size of the martensite (see page 233) is maintained throughout ordinary tempering operations. It is important to note that these grains are essentially B.C.C. ferrite in which the carbide particles are embedded.

Examining Figs. 113 to 118 more closely, the first change in appearance of the martensite is a general darkening as the tempering temperature is increased. This is probably caused by formation of Fe_3C particles and their growth to a critical size which etches rapidly (Fig. 115). Further coagulation of cementite at higher temperatures finally gives particles of such a size that they are discernible as such at X500. Meanwhile the white ferritic matrix becomes more prominent. As pointed out on page 165, both ferrite and cementite remain white after etching if they are in massive form. According to the definition of tempering (page 203), reheating is followed by any desired rate of cooling. Normally this cooling rate has no effect on the properties of the tempered steel; however, certain grades of alloy steels are susceptible to " temper brittleness " upon cooling slowly and should, therefore, be quenched from the tempering heat for highest impact toughness. (See pages 299 and 488.)

Properties of Tempered Steels. — While only the hardness values are available for the set of tempered steel structures of Figs. 113 to 118, it is a well known fact that tensile strength decreases along with hardness, while ductility increases as the tempering temperature is raised. Typical relations between the mechanical properties of heat treated steels will be found in Chapter IX. In general it may be said that the combination of strength, wear resistance, toughness, and ductility is high for structures such as Figs. 115 and 116.

Temper Colors. — These polished and etched microstructures should not be confused with the temper colors which are visible on the surface of roughly polished steel after tempering or drawing within certain temperature limits. Temper colors are a measure of surface oxidation. Very thin oxide films are pale yellow or straw colored while somewhat thicker films are blue and finally black. These colors may be produced on any steel, soft annealed or hard-

ened, by heating to the proper temperature for a suitable time. The proper drawing range for many tool steels and carburized parts coincides with the temperatures at which temper colors form, consequently the colors may be used for reasonably close control of the drawing of these steels when pyrometers are not available. Temper colors corresponding to various tempering or drawing temperatures depend to a large extent upon the time at temperature, chemical composition, and other variables. The following data are representative of short heating times at the indicated temperatures.

425° F.	pale yellow
470°	straw yellow
490°	brown yellow
530°	purple
550°	pale blue
600°	dark blue

It should be noted that the temperature range covered is below that in which marked changes in the microstructure occur.

Spheroidized Steel. — The tempered martensite structures shown in Figs. 113 to 115 may be arbitrarily classified as fine sorbitic, and those of Figs. 117 and 118 as coarse sorbitic. Long heating periods at temperatures just below or sometimes alternately above and below the Ac_1 line produce marked coarsening of the Fe_3C particles as in Fig. 119, and the product is said to have a spheroidized structure. For some purposes spheroidized steel is more readily machined than soft annealed steel because the hard carbide particles present are embedded in a continuous groundmass of soft, plastic iron. In certain cases, however, the spheroidized structure is actually too plastic for best machining and a laminated pearlitic structure or a combination of both gives better results. The most desirable structure will vary with the carbon content and the type of machining to be done.

Nomenclature of the Microconstituents in Steel. — The terms used to designate the microconstituents of steel are, at present, in a period of change. The terminology originally adopted has been found to be incongruous and inadequate in the light of modern knowledge of the nature of the microconstituents. Although a simple and more suitable nomenclature is coming into quite general

use, it has not been accepted by all authorities in the field and the entire subject is open to debate as this is written.[8, 11]   The newly proposed terms have been arbitrarily selected for use in this book.

The principal point of divergence between the two systems is in naming the dark etching structures formed on cooling at various rates lower than the critical cooling rate.   All of these structures consist of cementite particles of various degrees of fineness embedded in a matrix of ferrite, and according to the proposed terminology they are given the group name *pearlite*.   The cementite particles become increasingly fine as the cooling rate approaches the critical cooling rate, and the dark microconstituent often found associated with martensite (see Figs. 110 and 111) is now being called *fine pearlite*. This microconstituent has long been known as *troostite* and it is quite possible that this term will be retained.   The following comparison has been made in order to clarify the relationship of the old and newly proposed terms.

Proposed Term	*Origin and Description*	*Old Term*
Pearlite	Formed on cooling slowly as in annealing. Its composition is that of the eutectoid, approximately 0.83% carbon for plain carbon steels. The eutectoidal ferrite and cementite often form a coarsely laminated structure. (See Figs. 88 to 95.)	Pearlite (or lamellar pearlite)
Pearlite	Usually formed at rates intermediate between slow cooling and quenching (as in normalizing). Structure appears to be unresolved or very finely laminated at moderate magnification, and definitely lamellar when resolved at high magnification. May contain somewhat less than the eutectoid carbon content. (See Fig. 109.)	Sorbite
Fine pearlite *	Formed at cooling rates approaching the critical cooling rate, hence often found associated with martensite. Black structure unresolved * at X1000 but resolvable into a lamellar structure at very high	Troostite

* The adjective " fine " in the proposed term refers to the size of the carbide particles (or lamellae) within a grain of pearlite and not to the size of the pearlite grain itself.   It is entirely possible for " fine pearlite " to have a coarse grain size.

Proposed Term	Origin and Description	Old Term
	magnification. May contain less than the eutectoid carbon content. (See Figs. 110, 111, and 128.)	
Martensite	Light etching structure formed at cooling rates equal to or higher than the critical cooling rate. (See Figs. 106, 112, and 131.)	Martensite
Tempered martensite or Sorbite (or fine sorbite)	Dark etching microconstituent formed from martensite at relatively low tempering temperatures. Consists of a very fine dispersion of carbide particles in ferrite. (See Figs. 114 and 115.)	Troostite
Tempered martensite or Sorbite (or coarse sorbite)	Formed by tempering martensite at relatively high temperatures. Spheroidal cementite particles embedded in ferrite are visible at moderate magnification. (See Figs. 117 and 118.)	Sorbite
Spheroidized structure or Spheroidite	Formed by prolonged heating at high tempering temperatures. Consists of coarse spheroidal cementite particles in ferrite. (See Fig. 119.)	Spheroidized structure
Bainite	Obtained by direct transformation of austenite at a constant elevated temperature above the Ar″ transformation. (See Figs. 124 and 125.)	——

The principal objection to the old nomenclature is the dual use of the terms troostite and sorbite. Bainite is the newest addition to the list; however, it seems destined to become the proper term to apply to certain dark etching transformation products formed under special conditions.

Properties of Pearlitic versus Sorbitic Structures. — Examination of the foregoing data for the 0.72% carbon steel will show that there is considerable overlapping in the hardness values obtainable in pearlitic structures (produced by cooling at various rates lower than the critical cooling rate) and in sorbitic structures (produced by drastic quenching followed by tempering). It is true that a steel suitable for a given application may sometimes be heat treated by

either of these two methods to obtain one or the other of these structures; however, the quenched and tempered (sorbitic) steels ordinarily have superior tensile properties for a given hardness, as shown in Fig. 120. Only the ultimate tensile strength, computed on the original cross section, was found to be greater for the pearlitic steel.[8]
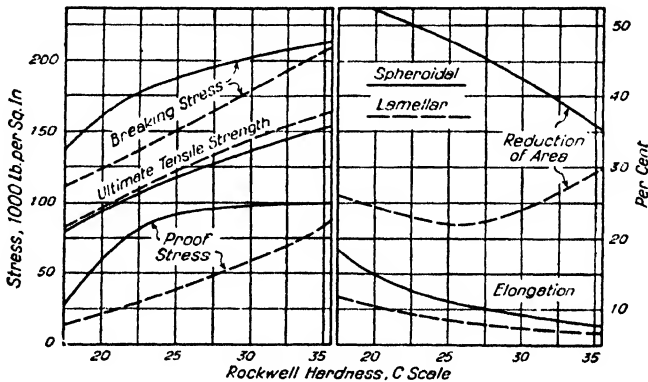
FIG. 120. — A COMPARISON OF THE TENSILE PROPERTIES OF SORBITIC (SPHEROIDAL) STRUCTURES OBTAINED BY TEMPERING MARTENSITE, AND PEARLITIC (LAMELLAR) STRUCTURES OBTAINED DIRECTLY BY COOLING AT VARIOUS RATES. THE SAME STEEL WAS USED FOR ALL TREATMENTS AND THE COMPARISON IS MADE AT VARIOUS HARDNESS VALUES. THE BREAKING STRESS IS COMPUTED ON THE REDUCED SECTION AFTER FRACTURE. SEE PAGE 496 FOR METHOD OF DETERMINING PROOF STRESS. (*Courtesy of J. R. Vilella, G. E. Guellich, and E. C. Bain, Trans. Am. Soc. for Metals.*[8])

In the above example, Fig. 120, pearlitic structures were obtained ranging in hardness from Rockwell " C " 17.5 (annealed) to Rockwell " C " 35.5 (produced by a cooling rate between C and B of Fig. 107). It will be recalled that hardness values of Rockwell " C " 41 and 63 were obtained by using quenching rates slightly lower than the critical cooling rate. (See Figs. 110 and 111.) Split transformation structures obtained in this manner have hardness values covering the entire range from the hardness of fine pearlite to that of martensite. Grossmann [12] has shown that the hardness increases, in this range, almost linearly with the martensite content, as shown in Fig. 121. Comparative tensile data are not available for these structures versus the tempered martensite (sorbitic) structures of the same hardness (Figs. 113 to 115); however, the tempered struc-

tures are probably superior in toughness.   Moreover, the split trans-
formation structures cannot be substituted for tempered structures
because of the difficulty of producing the former throughout a mass
of steel.   The critical nature of the cooling rate required to give a
50% pearlite-50% martensite structure, for example, makes it diffi-
cult to reproduce even in a very small sample.

In summarizing this comparison it may be said that wherever it
is possible to harden a piece of steel by cooling it faster than its
critical cooling rate, it will be possible to obtain any desired hard-
ness from full hard to dead soft by varying the tempering tempera-
ture.   The structure pro-
duced by this method will
be quite uniform through-
out the section, will be sub-
ject to close control, and
will have superior proper-
ties compared with equally
hard structures produced
by direct cooling at various
uniform rates.   In the pre-
ceding sentence provision
has been made for further
consideration of structures

FIG. 121. — ROCKWELL HARDNESS VERSUS
PROPORTION OF FINE PEARLITE IN A MICRO-
STRUCTURE OF MIXED PEARLITE AND MARTENS-
ITE.   (*Courtesy of M. A. Grossmann, Metal
Progress.*[12])

obtained by stepped cooling procedures or by cooling at nonuniform
rates.   (See page 221.)

Maximum Hardness. — Burns, Moore, and Archer [13] have shown
that the maximum attainable hardness of the common S.A.E. steels
of both carbon and low alloy grades varies with the carbon content
as shown in Fig. 122.   It appears that carbon content is more signifi-
cant than moderate amounts of alloying elements, grain size, or
other factors in determining the maximum attainable hardness.
However, in order to develop maximum hardness it may be necessary
to use very drastic quenching operations for certain steels.

Definition of Hardenability. — The term hardenability will be
used to describe the relative ease with which a steel may be fully
hardened rather than its maximum hardness.   This is often deter-
mined as a function of the depth to which a given sample may be

hardened by a given quenching procedure. On this basis harden-
ability is inversely proportional to the critical cooling rate, a high
hardenability corresponding to a low critical cooling rate. Critical
cooling rates will now be reconsidered from the standpoint of the
" S " curve.

Bain's " S " Curve. — The chart showing the effect of cooling rate
on the beginning of transformation of a eutectoidal steel, Fig. 107,
is actually an outgrowth of experimental work on transformation at

various constant tempera-
tures, as carried out by Bain
and his associates.[5 to 10]
These experiments are con-
veniently summarized in
the form of " S " curves
such as Fig. 123[5], which
was the first to be worked
out. Similar curves have
since been developed for
many carbon and alloy
steels. In practically all
cases the general shape of
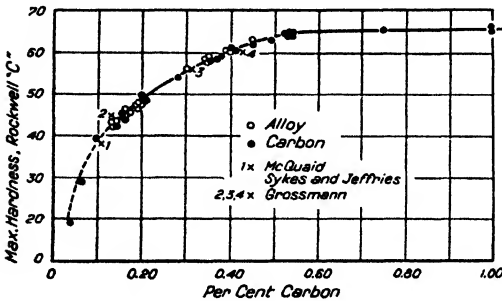
FIG. 122. — MAXIMUM HARDNESS VERSUS
CARBON CONTENT. (*Courtesy of J. L. Burns,
T. L. Moore, and R. S. Archer, Trans. Am. Soc.
for Metals.*[13])

the curve is similar to Fig. 123, although the time values, and to a
lesser extent the temperatures, may vary considerably depending
upon the characteristics of the steel being investigated.

In order to bring the steel to a low temperature, say 600° F., with-
out the formation of fine pearlite in the range of rapid transforma-
tion near 1000° F., it is necessary to cool very rapidly. It was found
that by using sheets 0.060 inch thick and quenching into molten lead
alloys * held at the desired temperature, such as 600° F., it was pos-
sible to prevent transformation of the austenite down to the holding
temperature, whereupon the specimen could be maintained at this
constant temperature any desired length of time, or until transfor-
mation of austenite was complete. At suitable time intervals speci-
mens were removed from the constant temperature bath and .
quenched in water to transform all retained austenite to the typical
white martensitic structure.

* Water was used as the holding bath to effect transformation at 160° F.
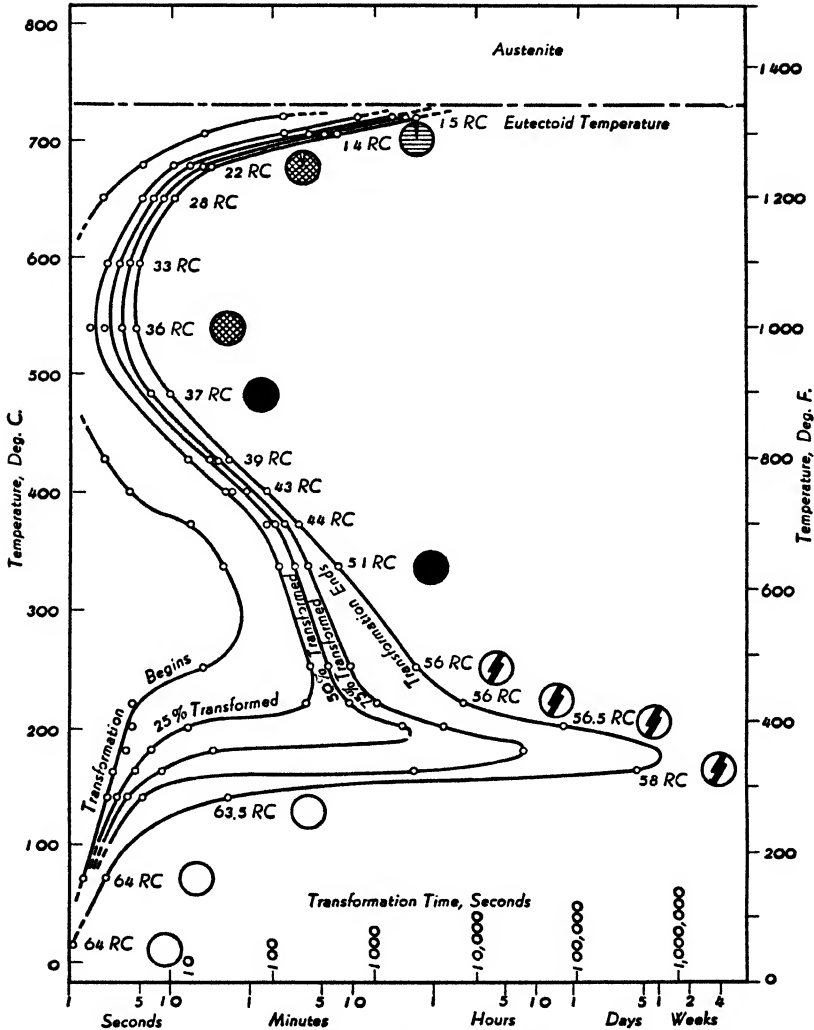
FIG. 123. — "S" CURVES SHOWING RATE OF DECOMPOSITION OF THE AUSTENITE AT CONSTANT SUBCRITICAL TEMPERATURES IN A STEEL CONTAINING 0.78% CARBON, 0.36% MANGANESE, AND 0.16% SILICON. (*Courtesy of E. S. Davenport and E. C. Bain, The Alloys of Iron and Carbon, Vol. I — Constitution.*[2] *Originally published in Trans. Am. Inst. Mining and Met. Eng.*[5])

Since the products of transformation at the holding temperature are dark etching, it was possible to estimate rather accurately the relative proportion of the structure which had undergone transformation at the elevated holding temperature before quenching in water. When these data were plotted, it became evident that there were two temperature ranges of rapid and two of slow transformation. In Fig. 123, for example, austenite transforms very slowly to pearlite at a temperature slightly below the equilibrium eutectoid transformation of 1335° F. It also transforms very slowly to a structure called bainite at about 350° F. On the other hand, austenite transforms very rapidly to fine pearlite at temperatures between 1100° and 1000° F., and to martensite at temperatures below 200° F.

The physical reasons for these changes in transformation rates with temperature, and in the products obtained, have been considered by Mehl [14] who has succeeded in establishing a working hypothesis correlating the known data on the effect of grain size, rates of nucleation, and rates of growth during transformation to pearlite; rates of diffusion of carbon and alloying elements in austenite; effect of alloying elements on rates of growth; the rate and manner of formation of martensite; and other factors pertaining to the transformation of austenite. Reading this paper is undoubtedly one of the best procedures for studying the significance of the " S " curve.

Products Obtained by Constant Temperature Transformation of Austenite. — The pearlitic structures which form at approximately 1100° F. and at higher temperatures are practically the same whether the steel is cooled at a uniform rate or made to transform at constant temperature. However, as the critical cooling rate is approached there is the possibility of mixed pearlite plus martensite structures (see Figs. 110, 111, and 121) caused by partial transformation at the elevated temperature. At low transformation temperatures the martensitic structures obtained by either process are identical. At intermediate temperatures between about 300° and 1000° F. constant temperature transformation produces a new type of structure called " bainite." When allowed to transform completely in this range, the microstructures are dark etching and very

similar in appearance to fine pearlite or " troostite." By arresting the transformation before completion (quenching in water) the remaining austenite may be changed to white etching martensite which forms a suitable background for observing the structure of the bainite. The structures obtained in this way by partial transformation

Left. FIG. 124. — EUTECTOID STEEL PARTIALLY TRANSFORMED AT 850° F. THE WHITE CONSTITUENT IS MARTENSITE PRODUCED BY QUENCHING IN WATER BEFORE THE TRANSFORMATION WAS COMPLETE AT 850° F. (Etched with picral.) X1500. (*Courtesy of Harry Tobin.*)

Right. FIG. 125. — EUTECTOID STEEL PARTIALLY TRANSFORMED AT 500° F. (Etched with picral.) X1500.

at temperatures below the 1000° F. hump of the " S " curve may have the characteristic feather-like appearance of Fig. 124 (sometimes called " high bainite "), while at somewhat lower temperatures the transformation products are definitely acicular as shown in Fig. 125 (sometimes called "low bainite "). In both cases the structures are believed to consist of highly dispersed cementite particles in ferrite, somewhat similar to the sorbitic structures obtained by quenching to martensite and tempering.

Hot-Quenching of Steel. — " Austempering." — The hardness values for a eutectoid steel transformed at various constant tem-

peratures are given in Fig. 123. Comparison with Figs. 113 to 116 shows that transformation at temperatures between 400° and 1000° F. gives hardness values approximately equivalent to those obtained by quenching followed by tempering in this same temperature range. It has been found that quenching in a molten lead alloy bath or fused salt bath and allowing transformation of austenite to go substantially to completion at an elevated temperature before quenching in water gives mechanical properties which are highly satisfactory for many applications.* Compared with ordinary quenching and tempering practice, " hot-quenching " usually gives greater ductility and impact toughness for a given hardness, and greatly reduced possibility of warping or cracking during hardening. The advantages of hot-quenching are most evident at the higher hardnesses corresponding to low tempering temperatures. The comparison made in Table XXIX shows the favorable mechanical properties of a hot-quenched steel containing 0.74% C, 0.37% Mn, 0.145% Si, 0.039% S, 0.044% P.

The principal limitations of the process are: inability to obtain very high hardness (over 60 Rockwell " C ") except by direct transformation to martensite at temperatures below the lower hump of the " S " curve, and the difficulty of cooling below the 1000° F. hump of the curve without transformation to soft products. Unless the sections to be hardened are very thin, it is necessary to use a steel with a low critical cooling rate (one whose " S " curve is shifted to the right on the time scale) in order to obtain uniform results throughout the section.

It has recently been shown by Payson and Hodapp[15] that in the case of most S.A.E. alloy steels the Charpy impact toughness obtained by normal quenching and tempering treatments is greater than that obtained by austempering for hardnesses less than about 42 R " C."

Microstructures Obtained upon Cooling a Hypoeutectoid Steel. — For simplicity nearly all the steels discussed in this chapter have been eutectoid or near eutectoid types. Actually the hypoeutectoid S.A.E. steels are at least as important as the higher carbon steels *from the standpoint of hardening practice.*

* This process, also known as " austempering," is patented.[7]

TABLE XXIX.   COMPARISON OF MECHANICAL PROPERTIES OF HOT-
QUENCHED VERSUS QUENCHED AND TEMPERED 0.74% CARBON
STEEL

	Hot Quenched	Quenched and Tempered
Rockwell " C "	50.4	50.2
Yield point, lb./sq. in.	151,300	121,700
Tensile strength, lb./sq. in.	282,700	246,700
Elongation in 6 inches, %	1.9	0.3
Reduction of area, %	34.5	0.7
Impact, ft.-lb.*	35.3	2.9

* To break 0.180 inch round, unnotched specimen.

Hot-Quench Treatment: Quench from 1450° F. into lead alloy bath at 580° F., hold at 580° F. for 15 minutes, quench into water.

Quench and Temper Treatment: Quench from 1450° F. into oil at 70° F., temper for 30 minutes in lead alloy bath at 600° F., quench in oil.

Courtesy of E. S. Davenport, E. L. Roff, and E. C. Bain, Trans. Am. Soc. for Metals.[7]

Only after relatively slow cooling or upon transformation at temperatures over the 1000° F. hump of the " S " curve do the microstructures of hypoeutectoid steels differ materially from those already observed.   The presence of free ferrite after these treatments distinguishes the hypoeutectoid from the eutectoid steels, as will be illustrated in the following heat treatments of a steel containing 0.50% C, 0.71% Mn, and 0.22% Si.   Samples ⅞ inch in diameter were heated to 1560° F. and cooled at various rates as indicated in Table XXX.   (The microstructures and hardness tests were taken midway between the surface and center.)

The slowly cooled structure, Fig. 126, shows the result of normal transformation of austenite in the critical range between Ar_3 and Ar_1.   (See Figs. 84 to 89.)   Faster cooling tends to suppress the formation of free ferrite, increase the fineness of the pearlite, increase the proportion of pearlite to ferrite, and increase the hardness and strength.   Quenching in a bath held at 300° F. gave a structure, Fig. 127, containing a relatively small proportion of ferrite, while oil quenching, Fig. 128, prevented the formation of massive ferrite units.   The carbon content of the pearlite has decreased

from about 0.80% in Fig. 126 to 0.50% in Fig. 128; nevertheless, the elimination of the massive ferrite regions and the increased fineness of the pearlite have resulted in an increase in hardness.

The microstructures produced by quenching at higher rates are practically the same as for higher carbon steels. (See Figs. 110 to 112 for example.) It is practically impossible to distinguish be-

TABLE XXX.   HEAT TREATMENT OF A 0.50% CARBON STEEL

Treatment	Structure	Hardness
A Cooled with the furnace (Annealed)	Ferrite and laminated pearlite (Fig. 126)	Rockwell " B " 92
B Cooled in still air (Normalized)	Ferrite and pearlite (pearlite unresolved at X500)	Rockwell " B " 95
C Quenched in a fused salt bath held at 300° F.	Smaller amount of ferrite and fine pearlite (Fig. 127)	Rockwell " B " 99 Rockwell " C " 21
D Quenched in oil at room temperature	Fine pearlite (Fig. 128)	Rockwell " C " 25
E Quenched in cold water without agitation	50% fine pearlite 50% martensite	Rockwell " C " 45
F Quenched in cold water	Martensite	Rockwell " C " 61

tween carbon contents from 0.15% up to about 0.90% by examination of the microstructures of fully hardened steels. Hypereutectoid steels may, however, be distinguished by the presence of excess cementite after hardening. (See Fig. 245, page 520.) Annealed structures are far more satisfactory for estimating carbon content.

Special Quenching Procedures. — An old heat treating procedure known as time-quenching is somewhat related to hot-quenching in that it takes advantage of the relatively slow transformation of austenite once the hump of the " S " curve is passed. Properly heated steel is quenched in water for a definite period of time or until the temperature is well below the zone of rapid transformation near 1000° F., then it is removed from the water and cooled to room temperature at a much lower rate in an oil bath. In this way the danger of cracking is materially decreased because the volume
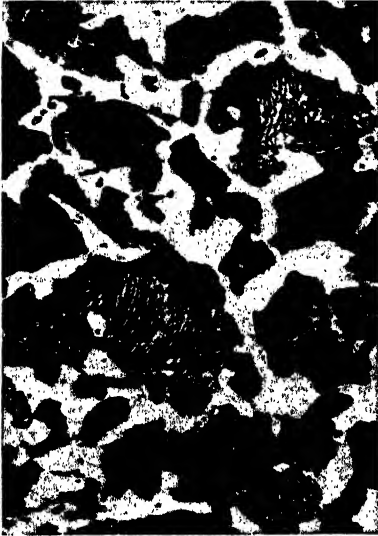
Left. Fig. 126. — 0.50% Carbon Steel Cooled Slowly from 1560° F. (Etched with picral.) Rockwell "B" 92. X500.

Right. Fig. 127. — 0.50% Carbon Steel Quenched from 1560° F. into a Fused Salt Bath Held at 300° F. (Etched with picral.) Rockwell "B" 99 -- Rockwell "C" 21. X500.
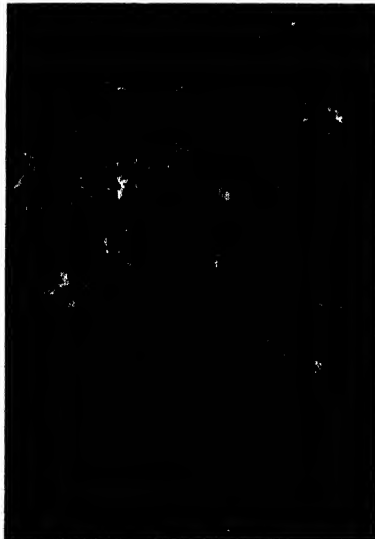
Fig. 128. — 0.50% Carbon Steel Quenched from 1560° F. into Oil at Room Temperature. (Etched with picral.) Rockwell "C" 25. X500.

increase accompanying martensite formation takes place at a lower rate, which tends to reduce the internal stresses. Another procedure which accomplishes the same result includes quenching in water until the drawing temperature is reached, transferring to a drawing furnace or liquid media, holding at the drawing temperature a definite length of time, then cooling either slow or fast depending on the size and shape of the piece.

Patent annealing, a treatment generally applied to wire during processing, is another form of quenching involving the use of hot coolants such as molten lead and fused salts. In this case the object is to cool the wire at a high enough rate to obtain a fine pearlitic structure, and, in the case of hypoeutectoid steels, to prevent the separation of free ferrite. Martensite cannot form because the quenching bath temperatures are too high — higher even than in the hot-quenching processes designed to give bainite. The uniform pearlitic structure obtained by patent annealing has been found to be highly suitable for subsequent wire drawing operations and the cold drawn product develops the high strength required for wire rope, suspension bridge cables, piano wire, etc. Photomicrographs of cold drawn wire were shown in Figs. 104 and 105, page 198.

Normalizing and Annealing. — The major portion of this chapter has been devoted to heat treatments involving quenching or special cooling procedures. Ordinary normalizing and annealing treatments (defined on page 202) are equally important. The results obtained by the latter treatments are not difficult to interpret; in fact, the low cooling rates used in annealing treatments permit transformations to proceed, for all practical purposes, in the manner indicated by the equilibrium diagrams. In the case of steels having relatively low hardenability, or steels heat treated in heavy sections, air cooling or normalizing likewise yields results not far removed from equilibrium conditions.

The commercial applications of annealing as a softening and homogenizing treatment are well known. In the case of normalizing the emphasis might well be reversed since it is generally used as a means of homogenizing or equalizing the structural conditions throughout the piece and for relief of internal stresses, as well as for softening following hot or cold work or previous heat treatments.

Annealing gives the lowest hardness because the steel is usually held at a temperature over the critical range for a longer time than in normalizing, and the cooling rate is lower, both of which tend to coarsen the microconstituents.

A long soaking period aids homogenization of cast structures; however, shorter holding time at higher temperatures, often followed by air cooling (normalizing), may be equally effective. Normalizing and combinations of annealing and normalizing are in general use for the heat treatment of steel castings. (See Chapter XI.)

The temperature used in annealing and normalizing treatments may vary from that given in the definitions. For example, forgings are often normalized from more than 100° F. above their upper critical temperature in order to coarsen the grain size for improved machinability. (See page 241.) To insure maximum grain refinement both normalizing and annealing temperatures should be just above Ac_3.

In heat treating large castings and forgings it is often impractical to use liquid quenching baths. In normalizing and annealing heavy sections special factors are involved which are not ordinarily considered in heat treating smaller sections. The treatment given very large steel forgings used for steam turbine rotors, mill drives, steamship propeller shafts, reaction chambers, etc., is described by Schaefer [16]:

" The heat treatment usually begins by heating sufficiently high above the critical range long enough to effect a complete solution of the micro-constituents. Extreme care must be taken in this first heat, just as extreme care must be taken in the first forging operation. The majority of the strains left by the forging operation are relieved in this first heat.

" The piece should be charged into a heated furnace before it becomes cool from forging. It must be heated up to temperature very slowly and uniformly. It must be held at temperature a long time. The old rule of ' an hour per inch of diameter ' is a good one, even though the diameters are extremely large. Cooling from this first heat may take place preferably in the furnace or, on occasion, in the air.

" For grain refinement repeat this heat treatment either once or a
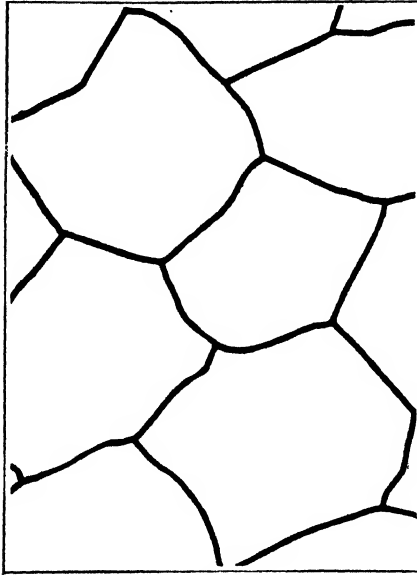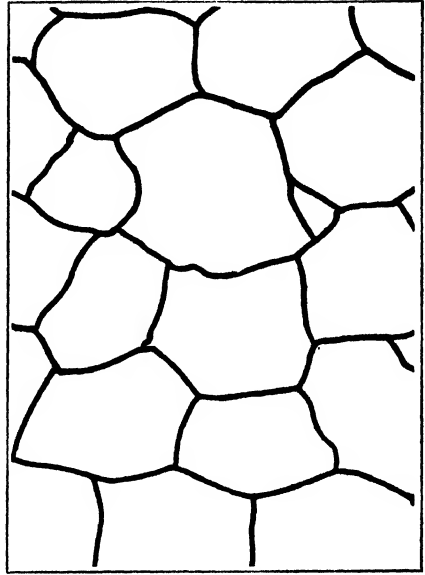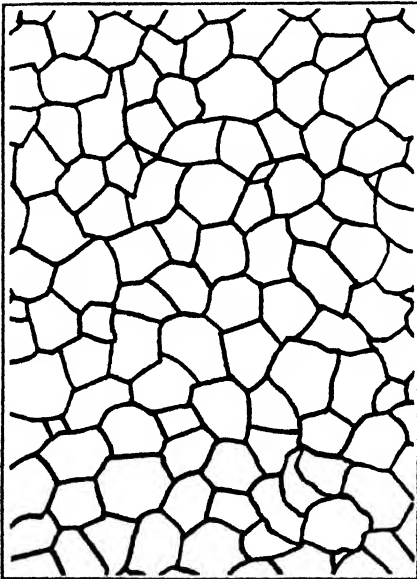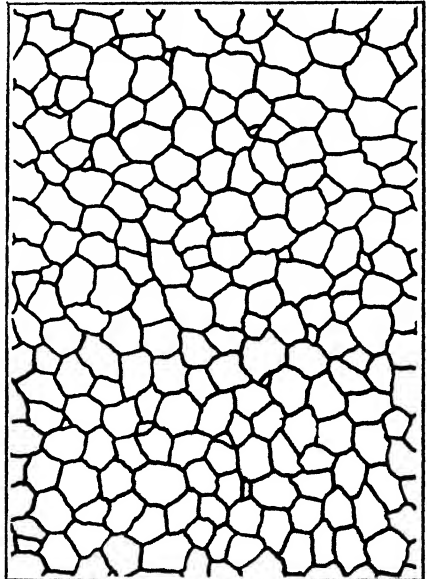
FIG. 129. — TENTATIVE GRAIN SIZE CHART FOR CLASSIFICATION OF STEELS.
Am. Soc. for Testing Mats.)
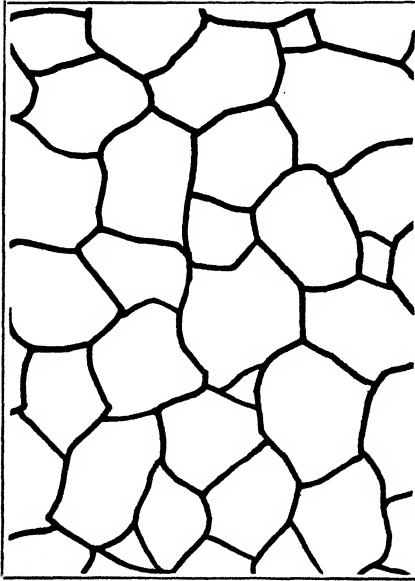
Grain Size No. 3
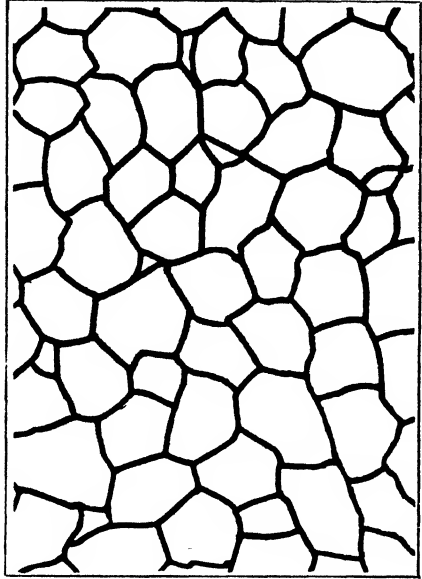
Grain Size No. 4
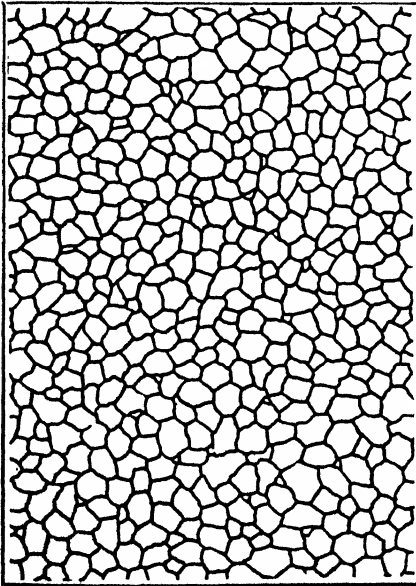
Grain Size No. 7

Grain Size No. 8

(A.S.T.M. Tentative Standard E19–38T.) Magnification X100. (*Courtesy of*
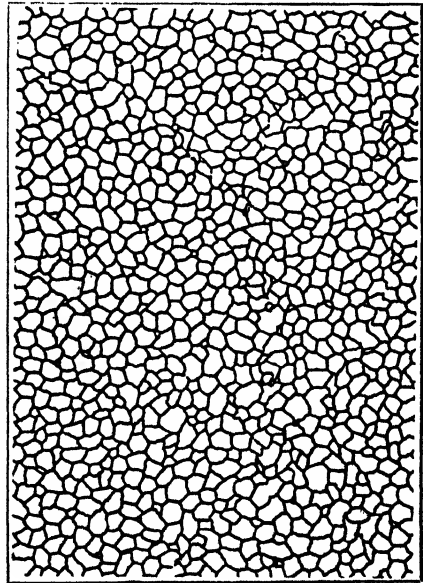
number of times at successively lower temperatures, each above the critical range. These heats are all performed slowly and carefully. In each case the forging is held at temperature approximately an hour per inch of cross-section, and the cooling is usually in air. Finally, heat below the critical range and cool slowly and carefully."

The practice described is satisfactory for carbon steel forgings. Some alloy steel forgings are more sensitive than carbon steels to thermal checks upon cooling after forging, therefore they are charged hot into a hot furnace, thoroughly soaked over the critical temperature range, and cooled very slowly to less than 400° F.[16]

Determination of Grain Size of Steels. — The general subject of the following portion of this chapter is the effect of composition and grain size on the hardenability of steel. Before proceeding, however, it will be desirable to study grain size as such and to become familiar with some of the methods for measuring hardenability.

Before the use of metallurgical microscopes became general, grain size was often determined by quenching, breaking, and observing the fracture. Recent investigations have proved that the simple fracture test does bear a close relationship to the microscopic grain size of hardened steels.[17] In fact, hardened steels generally break through the grain boundaries when fractured by impact (see Fig. 17, p. 19), and although the individual grains may be invisible to the unaided eye, the general fineness of the fracture may be readily evaluated on a comparative basis. In this country B. F. Shepherd [18] has worked out a standard set of fractures for use in his method for determining hardenability of tool steels. These sets are available by purchase and are being used to check the grain size of machine steels as well as tool steels.

The A.S.T.M. grain size charts [19] are used as the basis for several microscopic methods for determining grain size of steels. Actual photomicrographs are used in the 1933 A.S.T.M. standard, and the diagrammatic form reproduced in Fig. 129 has recently been adopted as a Tentative Standard (E19–38T). The information in Table XXXI, also from the Tentative Standard, gives the physical significance of the grain size numbers.

Although the A.S.T.M. grain size charts were originally applied

to hypereutectoid structures produced in the McQuaid-Ehn test (see page 234), they are now being used in determining the grain size of all types of steels using various methods of heat treatment. The following paragraphs illustrate some of the applications.

TABLE XXXI. GRAIN SIZE RELATIONSHIPS, ACTUAL AND AS OBSERVED AT 100 DIAMETERS MAGNIFICATION

A.S.T.M. Grain Size Number	Mean Number of Grains per Square Inch as Viewed at 100 Diameters	Calculated Diameter of Equivalent Spherical Grain (not magnified)		Calculated Mean Area of Cross-Section of Grain (not magnified), sq. in.
		in.	mm.	
No. 1	1	0.0113	0.287	0.000 1
No. 2	2	0.00800	0.203	0.000 05
No. 3	4	0.00567	0.144	0.000 025
No. 4	8	0.00400	0.101	0.000 012 5
No. 5	16	0.00283	0.0718	0.000 006 25
No. 6	32	0.00200	0.0507	0.000 003 13
No. 7	64	0.00142	0.0359	0.000 001 56
No. 8	128	0.00100	0.0254	0.000 000 78

Note. — Larger sizes Nos. 00 and 0 may be read by using a magnification of 50 diameters (instead of 100) and using grain sizes Nos. 1 and 2 on the chart, respectively. Smaller grain sizes Nos. 9 and 10 may be read by using a magnification of 200 diameters (instead of 100) and using grain sizes Nos. 7 and 8 on the chart, respectively.

[The original A.S.T.M. diagrams include black grains with white boundaries as well as white grains with black boundaries, and because of the larger field (at the same magnification) of these diagrams, it is recommended that they be used, rather than Fig. 129, where accurate estimation of grain size is required.]

Ferrite Grain Size. — The structure of rolled or annealed low carbon steel is predominately ferritic. When properly etched, the grain boundaries of the ferrite are clearly indicated and an A.S.T.M. grain size number may be assigned by comparison with the chart. For example, the grain size of the ferrite of Fig. 80, page 154, appears to be about No. 5 while that of Fig. 98, page 178 is No. 8. Small grains associated with much larger ones are neglected unless they appear in groups, as in Fig. 130, in which case two grain size numbers may be reported, as, for example: 50% No. 1–50% No. 7.

Austenite Grain Size. — It is often desirable to know the size of the austenitic grains as they existed at the annealing, normalizing,

or hardening temperature. Since austenite transforms to other products when carbon steels are cooled, the prior austenite grain size must be determined from the appearance of these transformation products.

Slowly cooled structures will be considered first. Grain size is readily estimated in fully annealed steels with carbon contents some-
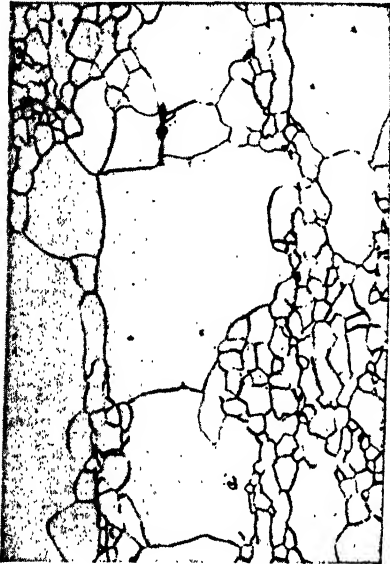
FIG. 130. — MIXED GRAIN SIZE IN AN IRON-SILICON ALLOY. (Etched with nital.) X100.

what lower or higher than the eutectoid composition. During cooling of these steels either ferrite or cementite separates first at austenite grain boundaries as in Figs. 90 and 95, thereby outlining the dark pearlite grains which form from the remaining austenite. Although each dark pearlitic unit in the microstructure may consist of several smaller grains having different orientations, each unit marks the location of a former austenite grain. In true eutectoid steels the white grain boundary network is absent, as in Fig. 91, and it is difficult to determine the prior austenitic grain size after slow cooling. The pearlite grains which are discernible in Fig. 91 because of their different orientations are probably smaller than the austenite grains from which they formed.

In medium carbon steels containing appreciable proportions of ferrite, as in Fig. 126, it is necessary to include one-half of the surrounding ferrite areas with each pearlitic unit in estimating prior austenitic grain size.

In low carbon steels and irons the ferrite grains which form, upon cooling, at nuclei within the austenite grains are indistinguishable from those which originate at the austenite grain boundaries, and the resulting ferrite grain size may be quite different from the prior austenite grain size and will, in general, be somewhat finer. More-

over, the pearlite associated with the ferrite grains in low carbon steels is generally not sufficient in amount to aid in demarcation of the former austenite grains. In order to overcome these difficulties a special carburizing procedure known as the McQuaid-Ehn test may be used to determine austenitic grain size of low carbon steels. (To be described in the next section.)

The rate of cooling of the austenite has no effect on the previously existing *austenite* grain size as measured by the above methods or those to follow. However, in the case of very low carbon steels and irons it has been observed that heating above the critical temperature range followed by rapid cooling, as in normalizing sheets, gives finer *ferrite* grain size than obtained upon cooling at somewhat lower rates. This may be due to the decreased time for grain growth of austenite when the time at which the steel is held above Ar_3 is decreased, or to transformation of austenite to ferrite at a greater number of nuclei upon rapid cooling. Experimental evidence favors the latter premise.

Higher carbon materials which

FIG. 131. — GRAIN SIZE OF A HARDENED EUTECTOID STEEL (MARTENSITIC STRUCTURE) REVEALED BY OXIDATION OF THE GRAIN BOUNDARIES AT THE SURFACE OF THE PIECE. (Etched with 15% HCl in ethyl alcohol.) X100. (*Courtesy of Harry Tobin.*)

form fully martensitic structures upon quenching retain the grain size of the austenite in their transformed structures. Although martensite grains may consist of an aggregate of submicroscopic crystals, as required in one of the theories for the hardness of quenched steels, they have definite microscopic grain boundaries indicating a normal " gross " grain size which is the same as the grain size of the austenite from which they formed. It is usually very difficult, however, to distinguish the martensite grain boundaries in fully hardened struc-

tures such as Figs. 106 and 112. Vilella [17] has developed an etching procedure which accentuates the orientation effects of the individual grains. Herasymenko [20] and Tobin and Kenyon [21] have shown that austenite grain boundaries are oxidized or decarburized more readily than the interior of the grains, thus establishing a grain boundary network of material which is more readily etched than the martensite grains which form from the austenite upon quenching. Only a thin surface layer is ordinarily affected, therefore very light grinding and polishing is necessary. A microstructure prepared in this manner is shown in Fig. 131.

Another procedure for determining the austenitic grain size of steels requires cooling at a rate slightly lower than the critical cooling rate, thereby producing a martensitic structure containing areas of dark etching fine pearlite, as in Fig. 111. Since the fine pearlite forms first at grain boundaries of the austenite, it outlines the former austenite grains in the transformed structure. A convenient method for attaining the special cooling rate required is to quench one end of a bar in cold water or brine, allowing the other end to cool in air. There will be a zone near the liquid line in which the martensite will be outlined by a network of fine pearlite.

The McQuaid-Ehn Test. — The McQuaid-Ehn test was developed in 1922 primarily for the purpose of measuring the hardenability of carburizing steels.[22] It has since been widely used to measure the so-called " inherent grain size " of a large variety of steels. The test consists of carburizing in solid carburizing compound for 8 hours at 1700° F. After cooling slowly from this temperature, the surface of the specimen will contain more than 0.90% carbon and will show either a " normal " carbide network structure as in Fig. 132, an " abnormal " structure as in Fig. 133, or some intermediate condition. In any case the grain size as estimated by comparison with the charts reproduced as Fig. 129 or with the original A.S.T.M. photomicrographs [19] is known as the McQuaid-Ehn grain size.

The terms " normal " and " abnormal " as used in this test are based on the characteristics of the microstructure of the hypereutectoid case. Coarse grain structures are usually " normal " and consist of rather finely laminated pearlite with well defined lines of

cementite in the grain boundaries. Fine grain steels tend to be " abnormal " in that the cementite coagulates into massive particles, both in the boundaries and within the pearlite grains. In extreme cases the pearlite does not appear as such since all the cementite coalesces into large particles in a ferrite matrix. The practical effect

Left. FIG. 132. — " NORMAL " HYPEREUTECTOID STRUCTURE PRODUCED IN THE McQUAID-EHN TEST. (Etched with picral.)   X100.

Right. FIG. 133. — " ABNORMAL " HYPEREUTECTOID STRUCTURE PRODUCED IN THE McQUAID-EHN TEST. (Etched with picral.)   X100.

of fine grain size " abnormal " structures on hardenability and other properties will be considered later in this chapter.

Because a steel remains fine at 1700° F. is no indication that it will not coarsen at 1750° F. or some higher temperature. Killed steels, especially those killed by additions of aluminum, generally have a rather definite coarsening temperature range which usually lies well above the upper critical temperature. Unkilled steels coarsen more uniformly as the temperature rises above the upper critical.

While the use of one arbitrary coarsening temperature has been criticized, and other objections have been raised to the use of this

test, it is, nevertheless, serving a useful purpose.   The tendency towards austenite grain coarsening at all temperatures above the critical range is unquestioned.   The McQuaid-Ehn test tells how much coarsening, if any, may be expected up to 1700° F., and since most heat treating operations, including carburizing, are carried out below 1700° F., the practical utility of the test is apparent.   Nevertheless, it has frequently been pointed out that coarsening in the McQuaid-Ehn test at 1700° F. does not necessarily indicate that a steel will become coarse grained upon heating to some lower heat treating temperature from 1400° to 1700° F.   In order to fit the test conditions more closely to actual operating conditions a Tentative Standard for Determination of Austenite Grain Size has recently been adopted by the A.S.T.M.[19]   Two alternative procedures are given : first, the McQuaid-Ehn test; second, the use of any of the other approved procedures for determining prior austenite grain size providing the heating temperature is not more than 50° F. above the temperature to be used in actual heat treating operations, and the time at temperature is not more than 50% greater than in actual heat treatment.

Grain Growth and Refinement. — Long before microscopic methods for studying grain structures became generally available, steel treaters discovered by observing fractures that steels could be refined by heat treatment, whereas most other metals, including the brasses and bronzes, could only be coarsened by heating.   Since fine grain size was readily associated with improved toughness, the capacity for grain refinement without hammering or rolling was recognized as an important characteristic of steel.

Many detailed studies of grain size characteristics of steels have been made in recent years, resulting in the following observations which are now quite generally accepted.

(1) Upon heating coarse grained steels through the critical range the grain structure changes to austenite and a definite refinement in size usually results.   Steels which are already fine grained will be refined little if any upon transformation to austenite.

(2) Upon further heating above the critical range the austenite grain structure tends to coarsen by normal grain growth.

(3) So-called fine grained steels are those which resist coarsening

upon heating well above the critical range. There is much difference of opinion as to what forces tend to prevent austenite grain growth. (See pages 240 and 269.)

(4) There is no tendency for refinement of austenite grain size upon cooling from the maximum temperature reached down to the critical range. Upon cooling through the critical temperature range the grain size of the products of transformation is closely related to that of the prior austenite grains; in fact, the pearlite or martensite grain sizes of medium and high carbon steels are essentially the same as the prior austenite grain size, while the ferrite grain size of low carbon steels is often somewhat smaller because of transformation from several nuclei within and at austenite grain boundaries. In a sense this is also true when pearlite is the principal product of transformation. It has already been pointed out that a single dark etching unit of pearlite may consist of several portions having different orientations. (See Fig. 132.) However, since definite boundaries are usually difficult to observe, the entire unit is often considered to be a pearlite " grain."

(5) As stated in (2) the application of heat tends to increase the size of austenite grains. The only possibility of refinement of austenite grains, once they have formed, is by the application of mechanical work. Rolling, forging, upsetting, pressing, and extruding are the principal hot working processes. Ordinarily the grain refining action is increased as the total reduction in section is increased, as the working temperature is lowered, and as the finishing temperature is lowered. The breakdown and refinement of the austenite grains is especially effective in certain hot pressing operations and in hammer forging.

It is important to note that grain refinement by heating through the critical temperature range (item (1)) and by mechanical work (item (5)) has no connection whatever with grain refinement by cold work and recrystallization. (See page 14.) In the latter process no phase changes are involved and, in the case of iron and steel, the recrystallizing temperatures used are well below the critical temperature range.

Grain Refinement by Heat Treatment. — The degree of refinement possible by heat treatment (item (1) above) has, at times,

TABLE XXXII. EFFECT OF HEAT TREATMENT ON THE GRAIN SIZE OF TEN S.A.E. STEELS

S.A.E.	As Received	Coarsened	First Anneal	Second Anneal	Third Anneal	Annealed After 1 Quench	Annealed After 3 Quenches	McQuaid-Ehn Grain Size Core [c]	McQuaid-Ehn Grain Size Case
		Ferrite grain size in A.S.T.M. numbers (Held at 1600° F. for 1 hour in all annealing and quenching treatments)							
1120	9	5–6 (1750° F.)	5–6	6	6	6–7	6–7	5	1 + 3 [f] Normal
X1315	8	6 (2025° F.)	6–7	7–8	8–9	8	9	7–8	8 Abnormal
X1315	8	6 (2200° F.)	7	8	9	7–8	8–9	7–8	8 Abnormal
2015	7	5 (1750° F.)	5–6	6	6	6	6–7	4–5	3–4 Normal
2315	8	6 (2025° F.)	7–8	8	8–9	8–9	9	6–7	8 Abnormal
2315	8	5 (2200° F.)	7–8	8	8–9	8	9	6–7	8 Abnormal
4615	7	5–6 (1750° F.)	7	7	7	7	7–8	5	3 + 6 [g] Normal
		Austenite grain size in A.S.T.M. numbers [a] (Held at 1500° F. for 1 hour in all annealing and quenching treatments)							
1050	6	2 (2200° F.)	5	6	6	5	6	6	7 Abnormal
X1340	7	5 (1750° F.)	5	5–6	5–6	5–6	6	6	6 Normal
3140	7	0 + 2 [d] (2200° F.)	5	6	6	6	[b]	2 + 5 [e]	1 + 6 [h] Normal
5140	6	3 (1750° F.)	4–5	5	5	4	5	4	4 Normal
6140	8–9	2 (2200° F.)	5	6	7	[b]	[b]	8	9 Abnormal

[a] As estimated from the pearlitic structures produced by slow cooling in the annealing treatment.
[b] Difficult to estimate grain size because of the highly dispersed condition of the ferrite and pearlite.
[c] Ferrite grain size is reported for the core of the low carbon steels.
[d] 25% "0" — 75% "2"
[e] 15% "2" — 85% "5"
[f] 20% "1" — 80% "3"
[g] 40% "3" — 60% "6"
[h] 30% "1" — 70% "6"

been overestimated, consequently experimental data are presented in Table XXXII in order to illustrate some of the limitations as well as the possibilities of this procedure.   The materials used include five S.A.E. carburizing steels and five S.A.E. medium carbon steels heat treated as short bars.   (See Table XXXVII, page 273, for type compositions of the alloy grades.)   These steels are all alloy quality forging bars made by killed steel practice.   The as-rolled bars were first given the McQuaid-Ehn test.   This treatment at 1700° F. coarsened the grain size of five of the steels.   Steels 4615 and 3140 were within their coarsening temperature range at 1700° F. since their structures were only partially coarsened, as indicated in the last column of Table XXXII.   Several samples of each steel were then packed in carburizing boxes and heated rapidly to coarsening temperatures ranging from 1750° F. to 2200° F. as indicated in the third column of the table, followed by slow cooling.   (The time at temperature was four hours for the 1750° F. treatment and 15 minutes at the higher temperatures.)

Coarsened samples were refined by heating over the critical temperature and cooling.   Results are shown for one, two, and three successive annealing treatments, as well as for one quench from above the critical temperature followed by annealing, and for three successive quenching treatments followed by annealing.

In the low carbon group it was not possible to coarsen the ferrite grain size beyond No. 5 even at high temperatures.   In the subsequent treatments the abnormal (McQuaid-Ehn test) steels underwent the greatest refinement, reaching No. 9 grain size.   The S.A.E. 1120 steel did not return to its original as-rolled grain size even after four reheatings (three quenches followed by annealing).

The austenitic grain sizes of the medium carbon steels were coarsened considerably by the preliminary treatments applied except in the case of X1340.   The greatest refinement occurred in the first reheating and only in the case of S.A.E. 1050 did the grain size become as fine as in the original hot rolled state.

Grain Growth Characteristics of Killed Steels. — The persistent fineness of grain often encountered in killed steels has been the subject of much research and speculation.   Powerful grain growth inhibiting agents are at work in a steel which retains a No. 8 grain

size after heating to 2025° F. (The austenitic grain size of the carburized case of S.A.E. X1315 was No. 8 after treatment at 2025° F. in the experiments reported above.) Such steels would, of course, be fine grained by McQuaid-Ehn standards after treatment at 1700° F. and they generally have " abnormal " structures similar to Fig. 133.

Difference of opinion exists as to what forces tend to prevent austenite grain growth in steels. Three different viewpoints may be stated briefly as follows. First, small particles of alumina, vanadium oxide, and other oxides, as well as small undissolved carbide particles may inhibit grain growth by *mechanical obstruction* at growth boundaries.[9, 23] Second, the presence of certain alloying elements, or their oxides or carbides, *dissolved* in the austenite may so stabilize the austenite grains that their growth tendencies are materially reduced. Third, aluminum or oxides of aluminum and also of iron may be in solution in the iron and tend to prevent or delay the solution and diffusion of Fe_3C and other carbides upon heating. This in turn would tend to promote coagulation and unequal distribution of the carbides and thereby delay the transformation of grain boundary ferrite to austenite, so that the untransformed ferrite would act as a shield preventing growth of the entrapped austenite grains. This theory, while not lacking in experimental justification, is somewhat involved and it is advisable to read the reports of McQuaid[24] and Morris and McQuaid[25] * for further details. Brophy and Parker[26] discount the possibility of pure aluminum, as such, producing abnormal fine grained steels in the absence of oxygen.

In any case it is well known that steels having nearly identical chemical analyses may have vastly different grain growth characteristics depending on the methods of melting and deoxidation employed in the steel making process, and that the addition of aluminum for deoxidation purposes tends to give fine grained steels.[27, 28] The effect of grain size on the properties of the steel will be considered further in the next section.

* Part of a symposium on hardenability published by the American Society for Metals in 1939. R. S. Archer has recently summarized these papers in Metal Progress, **35**, 33, 141, and 257 (1939).

Practical Significance of Grain Size. — It would not be worthwhile to make extensive studies of grain growth characteristics, methods of refining grain size, and methods of measuring grain size were it not for the influence of grain size on the properties of the steel. In the case of steels which are to be hardened by heat treatment, grain size is particularly important from the standpoint of hardenability, sensitivity to cracking, and tendency toward warping, as indicated by Bain and Vilella [17] in Table XXXIII.

TABLE XXXIII. EFFECTS OF PREVIOUSLY EXISTING AUSTENITE
GRAIN SIZE

Property	Trends in heat treated steel products from fine grained austenite — Nos. 5–8 (as compared with similar coarse grained steels)
Hardenability	Shallower hardening
Toughness at same hardness	Tougher
Distortion	Less distortion
Quenching cracks	Usually absent — less prevalent
Grinding cracks	Less susceptible
Internal stress	Lower
Retained austenite	Less

Property	Trends for annealed and normalized products having fine grain size — Nos. 5–8
Machinability	Inferior for rough machining
Machinability (special cases)	Better for fine finish
Formability (special cases)	Inferior

Courtesy of E. C. Bain and J. R. Vilella, Metals Handbook.[17]

It is apparent from Table XXXIII that coarse grained steels tend to be deep hardening and to be superior for certain rough machining and forming operations, otherwise the finer grained steels are generally preferred, especially from the standpoint of safety in heat treatment. The greater toughness of fine grained steels in both the hardened and unhardened conditions has been definitely proved experimentally.

The observed differences in the heat treating characteristics of fine versus coarse grained steels are generally associated with the effect of grain size on the critical cooling rate. It is believed that transformation of austenite takes place first at grain boundaries

and especially at intersections of several grains. Inclusions and undissolved carbide particles may also act as nuclei for transformation; however, assuming the number of independent particles of this nature to be constant, it is apparent that a fine grained austenite will have a much greater number of transformation centers than a corre-

	temp. °F.	G.S. (Mean)
A	1800	2
B	1700	3
C	1575	3.5
D	1450	4
E	1375	5

Carbon	0.74
Manganese	0.41
Silicon	0.14
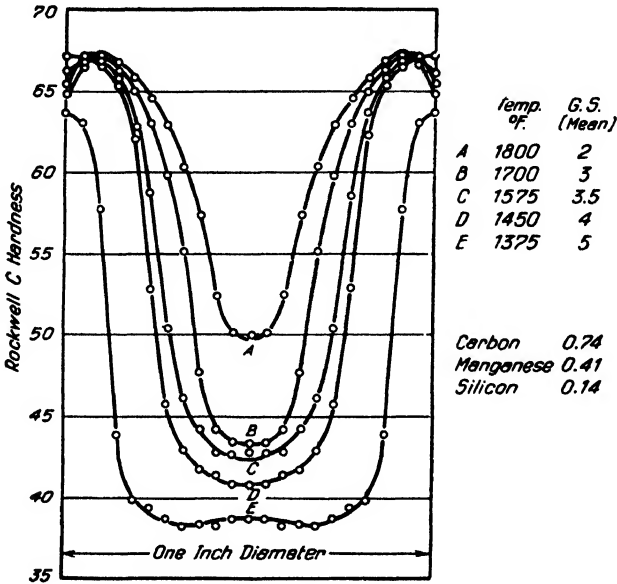
Fig. 134. — Hardenability of a 0.74% Carbon Steel as Influenced by Austenite Grain Size. Samples Were Brine Quenched from the Indicated Series of Temperatures. (Courtesy of E. C. Bain, Trans. Am. Soc. for Steel Treating.[29])

sponding coarse grained austenite, thus transformation will occur more readily upon cooling the fine grained steel. In order to prevent partial transformation to pearlite the quenching rate must be increased. In other words the critical cooling rate of fine grained steel, as defined on page 211, is higher than that of coarse grained steel, and higher quenching speeds are required to insure full hardness. Considering the " S " curve of Fig. 123, fine grain size causes a general shift to the left.

Bain illustrated the effect of grain size on the hardenability of a 0.74% carbon steel by heating to various temperatures, at which it developed austenite grain sizes from A.S.T.M. Nos. 2 to 5, then

quenching in brine.   The hardness characteristics obtained upon quenching are reproduced in Fig. 134.[29]

The transformation to martensite is generally accompanied by an increase in volume compared with the pearlitic structure of the unhardened steel.   Because of the volume changes accompanying this transformation and the fact that transformation cannot occur simultaneously through the entire piece because of its temperature gradient, it is easy to see that internal stresses of considerable magnitude may develop in the quenching operation.   It has been found that internal stresses are likely to be more severe in coarse grained, deep hardening steels, leading to greater possibility for cracking during or after quenching or in grinding operations.   It has already been stated that one of the objects of tempering is to relieve these stresses.

In high carbon and alloy steels there is a definite tendency to retain a part of the austenite untransformed upon quenching, thereby reducing somewhat the maximum hardness.   This is undesirable in the case of tools or carburized parts, for example, which may require high surface hardness.   Coarse grain size favors austenite retention.

The relationships between grain size requirements of steels for various automotive parts and such factors as chemical composition, section size, fabricating operations, and heat treating procedures are discussed by McQuaid.[30]

Measurement of Hardenability. — Variation in the response of steels to heat treatment has been the source of much difficulty for steel treaters.   In large scale production of heat treated machine parts it is very desirable to have stock with uniform hardening characteristics available, and in order to be sure that this is the case it may be necessary to apply some convenient test to determine the hardenability before the parts are processed.   Several procedures have been proposed, including the Shepherd P–F test, already mentioned, which depends on the appearance of specimens fractured after hardening at various temperatures, together with the depth of penetration of the hardened zone below the surface as observed on deeply etched cross sections.   (The full procedure will be found in a paper by Shepherd.[18])

Other methods involve Rockwell " C " hardness tests made on

cross sections of standard samples quenched under controlled conditions. The sections are cut far enough from the ends of the bar to eliminate end effects. Thin abrasive cut off wheels or nick and fracture methods are used to section the hardened steel. The hardness values are generally plotted against distance from the surface. (See Fig. 134.) The maximum hardness occurs at or near the surface while the minimum is at or near the center. (Retained austen-
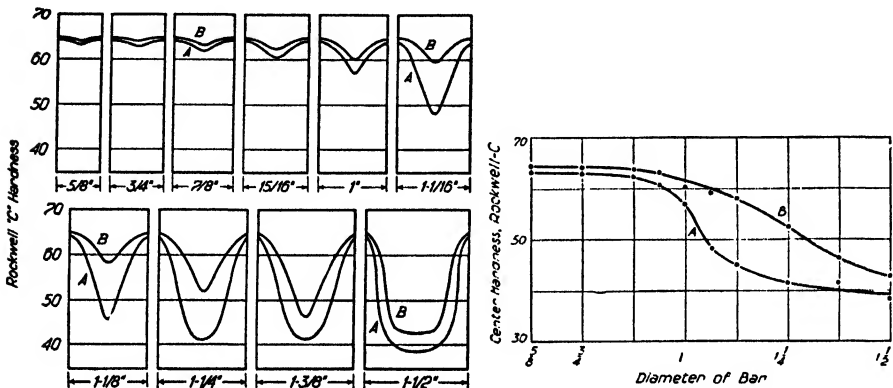
Left. FIG. 135.—HARDNESS DISTRIBUTION CURVES OF TWO STEELS DIFFERING SOMEWHAT IN HARDENABILITY. (*Courtesy of M. A. Grossmann, M. Asimow, and S. F. Urban, Trans. Am. Soc. for Metals.*[34])

Right. FIG. 136.—CENTER HARDNESS CURVES. DATA FROM FIG. 135.

ite or decarburization may reduce the skin hardness while segregation probably accounts for unusual center effects.) According to the procedure proposed by Burns, Moore, and Archer [13], the hardenability rating is expressed by the area under the hardness curve (based on zero Rockwell " C ") together with the surface and center hardness values.

A method proposed by Jominy and Boegehold [31] is particularly applicable to carburizing steels. Bars 1 inch in diameter by 3 inches are carburized at 1700° F. for 8 hours and quenched from the carburizing container, using a special fixture which confines the water spray to one end. Vickers-Brinell (10 kg. load) hardness tests are made along the length of the bar at the carburized surface and also at 0.015 inch below the surface. The distance from the quenched end of the bar within which the hardness is over 600

Vickers-Brinell is used as a criterion of hardenability. This method has also been applied to medium carbon steels.[32] It is claimed that the center hardness of bars up to 4 inches can be determined from the results of tests on 1 inch bars.

Test methods for determining the hardenability of tool steels are described by Schempp.[33] In addition to the Shepherd P–F test, a simple disc or slab test is used in which discs of graduated thickness are hardened and examined for soft core areas by the deep etching method. Schempp also describes a stepped bar test used for steels of eutectoid composition.

Effect of Section Size on Hardness Penetration. — The section size used in tests to compare the hardenability of various steels should be carefully considered. For each steel hardened under fixed conditions there is a range of critical section sizes in which the depth of hardness penetration changes rapidly with minor fluctuations in composition and quenching conditions. The effect of critical section size in comparing hardenability characteristics is illustrated by Grossmann, Asimow, and Urban[34] in Figs. 135 and 136. The points of inflection of the center hardness curves of Fig. 136 are taken as the critical section sizes for bars A and B (1 1/32 inch and 1¼ inch diameter respectively). The microstructures obtained in the centers of these critical sections are mixed martensite-fine pearlite structures having Rockwell hardnesses of approximately 50 to 55. (See also Fig. 121.) Grossmann, Asimow, and Urban[34] have made a detailed mathematical analysis of hardenability involving section size and severity of the quench. Luerssen[35], in reviewing the various methods for measuring hardenability, also considers the subject of critical sections.

Effect of Chemical Composition on Hardenability. — It has been stated that carbon content, alloy content, and grain size at the quenching temperature are the principal factors which control the critical cooling rate. High carbon content, the presence of alloying elements dissolved in the austenite, and coarse grain size at the quenching temperature all tend to lower the critical cooling rate, thereby increasing the hardenability.

The effect of carbon content on critical cooling rate has been determined by several investigators. The results obtained by

Digges [36], using high purity iron-carbon alloys, are summarized in Fig. 137. All alloys were quenched from 1700° F., at which temperature the austenitic grain size was A.S.T.M. No. 4 and all the carbon present was dissolved in the austenite. The critical cooling rate was taken as the lowest average cooling rate (measured between 1110° and 930° F.) capable of producing a martensitic structure with fine pearlite (troostite) present in amounts estimated to be between 1 and 3%.
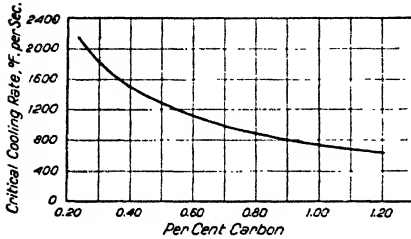
FIG. 137. — RELATION OF CRITICAL COOLING RATE TO CARBON CONTENT OF THE IRON-CARBON ALLOYS. (*Courtesy of T. G. Digges, Trans. Am. Soc. for Metals.*[36])

Using commercial steels, French and Klopsch [37] obtained somewhat lower critical cooling rates, reaching a minimum of about 300° F. per second at eutectoid carbon contents, and increasing with both lower and higher carbon contents. This increased critical cooling rate (reduced hardenability) of hypereutectoid steels, compared with eutectoid steels, is frequently observed in commercial heat treating operations. It has been explained by the fact that high carbon steels are generally quenched from a temperature above Ac_{1-3} but below Acm, consequently they contain undissolved cementite particles at the quenching temperature. The presence of particles of any kind such as undissolved carbides, oxides, or other inclusions increases the number of nuclei at which transformation may begin, thereby increasing the quenching speed necessary to prevent formation of pearlite, or shifting the " S " curve (see Fig. 123) to the left. Digges' results show that when the quenching temperature is high enough to insure solution of the carbides, the critical cooling rate decreases continuously with carbon content.

It has been found that certain carbide forming alloying elements such as vanadium and molybdenum tend to stabilize the carbides present and reduce their tendency to dissolve in the austenite at ordinary heat treating temperatures. Likewise certain elements such as aluminum and vanadium may be present as very finely divided oxides which are even less soluble in austenite. In either case such alloying elements may reduce the hardenability by introducing

transformation nuclei, even though it has been shown that they tend to increase hardenability provided they are fully dissolved in the austenite at the quenching temperature.   On page 240 it was pointed out that these same particles probably tend to obstruct grain growth of the austenite, thus fine grain size may contribute further to reduced hardenability.

In the case of aluminum the normal tendency is definitely towards fine grain size " abnormal " McQuaid-Ehn microstructures, and reduced hardenability, regardless of the basic cause of these related results.   A theory quite different from the fine grain size-nucleation theory is offered by Morris and McQuaid [25] who attribute the low hardenability of certain carbon steels to the presence of low carbon austenitic areas at the quenching temperature, resulting in separation of primary ferrite and a very high transformation rate in the 1000° F. region, thus preventing the formation of a fully martensitic structure even at the highest commercial cooling rates.   On this premise the effect of alloying elements on hardenability is explained as follows:

" Theoretically all that is necessary to prevent ferrite formation in the boundaries on quenching is to increase the temperature or time to the point where carbon diffusion across the boundaries had taken place and then to cool at a fast enough rate to retain the carbon in solution.

" Any element or combination of elements in the boundaries that will decrease the carbon solution will aid in the austenite transformation and decrease hardenability.   Conversely, any element which aids in the carbon solution or stabilizes the austenite formed will increase hardenability.

" We have seen that the pure iron-carbon combinations are inherently shallow hardening and the addition of many metallic elements will increase the hardenability.   Even aluminum in sufficient quantity decreases the critical cooling rate and hence increases the hardenability.   This is only true, however, when the temperature is sufficiently high and the time sufficiently long to insure complete carbon solution and uniform diffusion." [25]

Another interesting approach to the subject of hardenability of carbon steels has been made by Burns and Riegel [38], who deter-

mined the hardness-penetration characteristics of 1 inch diameter
bars quenched in water and were able to correlate the results with
chemical composition and grain size. Hardenability was measured
as the area under Rockwell " C " hardness-penetration curves simi-
lar to Figs. 134 and 135. The results obtained for steels in three
carbon ranges have been combined in Fig. 138, in which the ordi-
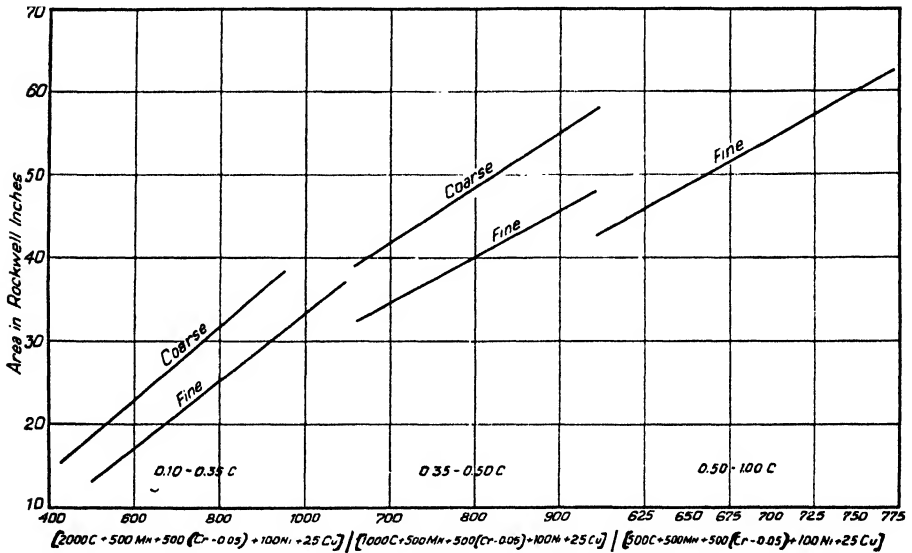
FIG. 138. — RELATIONSHIP BETWEEN CHEMICAL COMPOSITION AND AREA UNDER
THE HARDENABILITY CURVE FOR STEELS IN THREE CARBON RANGES. (*Courtesy of
John L. Burns and Glen C. Riegel, Trans. Am. Soc. for Metals.*[38])

nates are hardenability expressed in " Rockwell-inches," or area
under the hardness-penetration curve, and the abscissae are chemi-
cal composition factors in which each element present is weighted in
proportion to its estimated effect on hardenability. These factors
were selected by trial on the basis of establishing a linear relation-
ship between composition and hardenability. In order to maintain
this condition it was necessary to change the coefficient for carbon
content in the three carbon ranges. The grain size distinctions are
defined by Burns and Riegel as follows: " We are limiting the dis-
cussion here to killed plain carbon steels of a strictly fine or coarse
grain category (no intermediate grains) as represented by the grain
size at the quenching temperature."

It is possible to obtain from Fig. 138 some idea as to the relative importance of grain size and composition on the hardenability of low and medium carbon steels. (No data were available for coarse grained high carbon steels.) For example, the greatest effect of grain size on the medium carbon steels seems to be about 10 Rockwell-inch units or about 150 composition factor units. Additions of approximately 0.075% carbon or 0.30% manganese would apparently be sufficient to overcome the reduced hardenability of a fine grain compared with a coarse grained steel in the medium carbon range.

Silicon was present in these steels in amounts between 0.15 and 0.30%, and its effect on hardenability was considered to be small in this range. It should be noted further that chromium, nickel, and copper occur as residual elements in the charge rather than as specific additions in the class of steels studied. In the following chapter these elements will be considered in more detail.

Heat Treating Equipment and Practice. — Large scale heat treating operations are performed in batch or continuous type furnaces which are often equipped with automatic temperature regulators and mechanical devices for handling the steel in the furnace and in quenching baths, as well as means for regulating the furnace atmosphere to prevent scaling. Recent improvements in heat treating furnaces have been made possible because of the development of heat resisting alloys for moving parts, improved refractories, more efficient combustion units, electric resistor alloys for operation at higher temperatures, and innumerable high precision regulating devices. Unfortunately, space will not permit illustration or a detailed description of the various commercial units, nor have detailed recommendations for heat treating various grades of steel been included. For this information the reader is referred to such reliable sources as Metals Handbook, S.A.E. Handbook, A.S.T.M. Standards, and Alloys of Iron and Carbon — Vol. I. " The Relation of Design to Heat Treatment," Metals Handbook, 1939, page 214, is particularly recommended to engineering students.

Furnace Atmospheres. — When steel is heated in muffle type furnaces the atmosphere is normally oxidizing, and depending on the heating temperature, on the ratio of load to furnace volume, and on

the amount of leakage of air, the surface of the steel may be heavily coated or only slightly discolored by an oxide film.   By introducing a properly controlled atmosphere into the muffle or container the surface may be protected from oxidation.   Protection of the surface is not simply a matter of substituting a strongly reducing gas containing a high CO or hydrogen content, since decarburization may take place even though scaling is reduced or eliminated, giving low surface hardness if the parts are to be quenched.   On the other hand, slight decarburization may not be detrimental as, for example, in bright annealing of certain grades of low carbon steel sheets.   In general, the selection of a suitable atmosphere must be based on the composition of the steel, the purpose of the heating operation (rolling, forging, annealing, hardening, etc.), the temperature, and the type of furnace.

In the case of direct fuel fired furnaces the combustion products become the furnace atmosphere and it is usually more difficult to control the surface condition of the steel than it is in muffle type furnaces. Gas fired furnaces of the former type are naturally better adapted than oil or coal fired furnaces to atmospheric control.

Although rapid strides have been made in the past few years, the use of controlled atmospheres for heat treatment is still in a state of development.   The reader is referred to the numerous papers on the subject in the recent technical literature and to the 4th edition of Bullens' " Steel and Its Heat Treatment." [39]

Liquid Heating Media. — Molten lead and fused salt baths for hardening and tempering also have their place where freedom from scaling is important, especially in the heat treatment of tools and small parts.[40]   In addition to preventing scaling or surface decarburization, salt baths transmit heat quickly and uniformly compared with oven type furnaces, and when the heat is applied through submerged coils or electrical resistors the thermal efficiency is exceptionally high.[41]   Ordinary cyaniding as well as certain newly developed processes for surface hardening are carried out in molten salt baths.   While salt mixtures are available for heating high speed steels to 2350° F. for hardening, lead is seldom used over 1600° F. because of excessive volatilization.   However, lead is even faster heating than molten salt, gives complete protection to the surface,

and has certain other advantages. A detailed comparison of lead and salt baths for heat treatment will be found in Metals Handbook.[40]

Quenching Baths. — Having studied the significance of the critical cooling rate and its relation to hardenability, and considered the effect of cooling rate on the structure and properties of steels, it will be of interest to examine the various coolants from the standpoint of quenching speed and other characteristics.

Depending on the critical cooling rate and the section size and shape, steels may require a wide range of cooling rates for satisfactory hardening. Assuming that a certain coolant will give the proper cooling rate for a given treatment, it must satisfy certain other requirements including stability upon continued use, maintenance of its cooling characteristics within a reasonable range of bath temperatures, safety and convenience in handling, and low cost.

Scott [42] describes three stages of cooling in liquid quenching baths as follows:

" Stage ' A ' — *Vapor Blanket Cooling* — In this, the first stage, a thin stable vapor film surrounds the hot metal. Cooling is by conduction and radiation through the gaseous film and is therefore relatively slow.

" Stage ' B ' — *Vapor Transport Cooling* — Termination of the ' A ' stage is marked by wetting of the metal surface. Vapor forms copiously in bubbles and is carried away by gravity and convection currents to condense in cooler surrounding liquid. This is the fastest stage of cooling.

" Stage ' C ' — *Liquid Cooling* — The ' B ' stage ends as the surface temperature of the metal approaches the boiling point of the quenching liquid. Vapor no longer forms so cooling is by conduction and convection and the temperature difference is greatly decreased. This mode of cooling is slower than that of the ' B ' stage.

" Cooling in the ' B ' and ' C ' stages is in fair accord with Newton's law of cooling, that is the cooling rate is proportional to the surface temperature, which holds closely for short temperature ranges. In the ' A ' stage, however, cooling is at a nearly constant rate, which rate increases somewhat with increase in the bath temperature.

" Zavarine [43] has noted from high speed photographs that rate of bubble collapse is relatively slow in the ' A ' stage and extremely fast in the ' B ' stage, of course vanishing in the ' C ' stage. Also that crystals of salt are precipitated momentarily during quenching in brine."
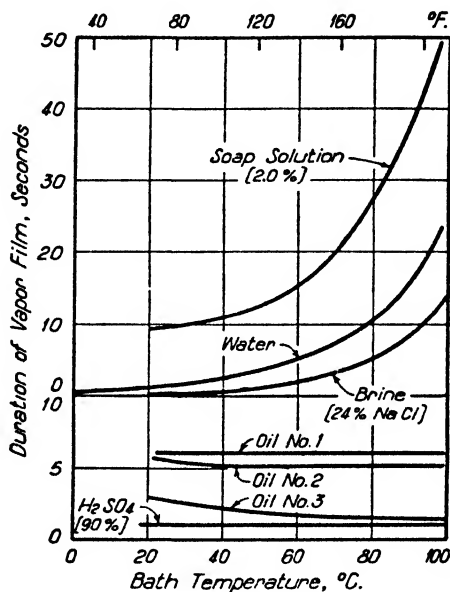
FIG. 139. — DURATION OF THE VAPOR FILM STAGE, "A," OF SURFACE COOLING DURING QUENCHING IN SEVERAL LIQUIDS. (*Courtesy of N. B. Pilling and T. D. Lynch* [44] *and Howard Scott* [42], *Metals Handbook.* [42])

" There has been much debate concerning the fundamental properties of liquids determining their quenching power. Those mentioned include : Thermal conductivity, viscosity, specific heat, heat of vaporization. Evidently which property is dominant depends on the stage concerned. Certainly boiling point is a major factor in the ' A ' stage, heat of vaporization in the ' B ' stage, and thermal conductivity in the ' C ' stage. Viscosity also contributes to cooling power both in the ' B ' and ' C ' stages. Each stage has an important part in determining the success of a quench."

The fundamental characteristics of the most common cooling media, water, brine, and oil, are indicated in the observations of Pilling and Lynch [44] in Fig. 139 and Table XXXIV.

TABLE XXXIV. COOLING RATES AT CENTER OF ¼ IN. DIAMETER
CYLINDER DURING QUENCHING IN STILL LIQUIDS AT
104° F. SHOWING VARIATION IN COOLING ACTION
WITH SURFACE TEMPERATURE OF PIECES

Quenching Liquid	Boiling Point, °F.	Duration of Vapor Film, Sec.	Relative Cooling Rate* in Stage: "A"	"B"	"C"
Tap water, up to 104° F.	212	3	0.10	1.00	0.16
Brine (24% NaCl)	221	1	—	1.00	0.15
Sulphuric acid (90%)	504	1	—	0.96	0.09
Soap solution (2%) up to 104° F.	—	11	0.08	—	0.15
Oil #1	712	6	0.16	0.36	0.035
Oil #2	756	5	0.17	0.48	0.042
Oil #3	860	2	—	0.27	0.027

Physical Properties of Oils

	#1	#2	#3
Flash point, ° F.	370	399	581
Fire point, ° F.	419	466	646
Boiling point, ° F.	712	756	860
Density	0.882	0.912	0.916

* Cooling rates in " A " stage taken at 1472–1292° F.; in " B " stage at 1022° F.;
and in " C " stage at 392° F.

Courtesy of N. B. Pilling and T. D. Lynch [44] and Howard Scott,[42] Metals Handbook.[42]

The relatively low cooling rate during the vapor blanket stage,
" A," is confirmed in Table XXXIV. The duration of stage " A "
at various bath temperatures is shown in Fig. 139. The well known
capacity of brine solutions (also acids and caustic solutions) for
" taking hold " quickly at the beginning of the quench is explained
by the shorter duration of the vapor film stage. The necessity for
keeping soap out of quenching baths is also apparent.

Water and water base quenching media are almost equally drastic
in the vapor transport or " B " stage. Fortunately the temperature
range of this stage includes the zone of rapid transformation to
pearlite at about 1000° F., making it possible to harden even the
shallow hardening steels provided the duration of the " A " stage
has been short enough to prevent transformation to pearlite at
higher temperatures.

Quenching oils have entirely different characteristics. The cool-

ing rate during the vapor film stage is higher than for water and the duration of this period is not unduly long, especially for oils with high boiling points. Heating of the bath does not increase the duration of the " A " stage as it does in the case of water and its solu-
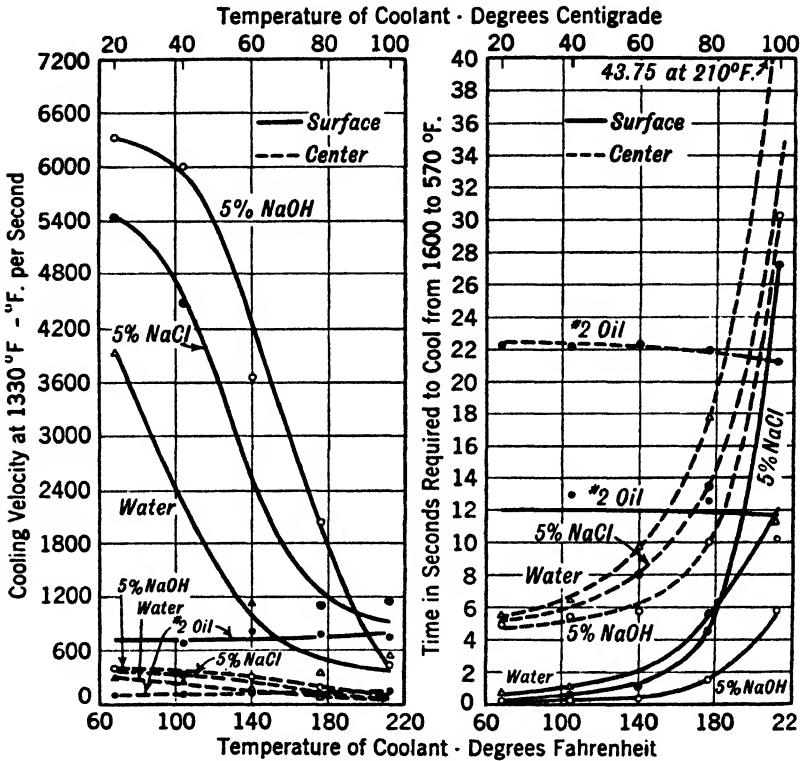
Fig. 140. — Effect of Temperature of Bath on Center and Surface Cooling Times and Rates in Different Quenching Liquids. (Courtesy of H. J. French, The Quenching of Steels.[45])

tions. Nevertheless, oil quenching is much less drastic than water because of the greatly reduced cooling rate in the " B " stage, hence only deep hardening steels having low critical cooling rates may be fully hardened in oil. The advantage of oil over water in hardening these steels is in the very low cooling rate of oil in the liquid cooling or " C " stage. It will be recalled that there is no danger of transformation of austenite to soft microconstituents in the lower portion of the " S " curve below about 500° F. Slow cooling in this

range during martensite formation, as in oil quenching, is conducive to relief of the stresses which are normally built up by the volume changes accompanying transformation, therefore greater toughness and less danger of quenching cracks is to be expected.

The physical properties and relative cooling rates of various commercial oils and a comparison of the cooling rates of oils with other coolants are given by Scott.[42]

Comparative cooling rate data obtained by French [45] for ½ inch diameter by 2 inch cylinders quenched into liquids at various temperatures are summarized in Fig. 140. The motion of the coolants was 3 feet per second. At room temperature the surface cooling rate for the quenching oil is very much lower than for water or dilute salt or dilute caustic solutions. The decline in cooling rate as the temperature of the bath increases is very rapid for water and its solutions, confirming the data of Fig. 139. It follows from this that proper circulation or cooling of water quenching baths is very important for continuous or large scale batch type heat treating operations. While it is true that the cooling characteristics of quenching oils are affected but little at bath temperatures up to 200° F., local temperatures may become much higher than this and there is the possibility of reaching the flash point when oils are overheated.

The danger of forming soft spots in water quenching due to the insulating effect of the vapor film is largely overcome when the water is circulated rapidly or used as a spray. Evidence of the well known fact that motionless or slowly moving water is much less effective than running water for hardening steel is given in Fig. 141. The formation of steam pockets and their adherence to the surface is responsible for the slow cooling in position 3, at which the cooling rate was even lower than in a motionless bath (curve 4).

Agitation of the steel is often desirable in still quenching baths provided the flow is uniform and more or less symmetrical about the work. Obviously the motion of the coolant in the above experiment would not be conducive to uniform hardness and would probably cause warping. The possibility of warping would be greatly reduced by agitation of the cylinder in a direction parallel to its axis in a motionless bath. Spray quenching and submerged spray quenching, involving the use of special fixtures for specific applications, are being

used very effectively for production parts such as gears and for the many forms of local hardening now being developed.

The advantage of using caustic solutions for drastic quenching is indicated in the lower section of Fig. 141.

Gases dissolved in quenching baths are liberated upon the introduction of hot steel. Together with steam, or the decomposition products of certain quenching media, they tend to form an insulating layer between the steel and the coolant. Carbon dioxide is one of the most undesirable of the gases which may be dissolved in water. It is even more soluble in caustic solutions and rather frequent renewal of these solutions is necessary for best results.

The scale which is present on unfinished portions of forgings or castings or which may be formed on the surfaces of articles heated in oxidizing atmospheres also acts as an insulating jacket and tends to confine the gases. A heavy scale which cracks off in quenching is less detrimental than an adherent scale. All factors which tend to remove these substances, or prevent them from forming, increase the cooling rate.

FIG. 141. — TEMPERATURE-TIME COOLING CURVES TAKEN AT THE SURFACES OF ½ INCH DIAMETER BY 2 INCH STEEL CYLINDERS QUENCHED IN A ROTATING BATH OF WATER OR 5% SODIUM HYDROXIDE. (*Courtesy of H. J. French, The Quenching of Steels.*[45])

Hardening by Means of Local Heating and Quenching. — This chapter will be concluded with a discussion of certain recently developed processes which are based on local rather than general heating. This principle is not new since gear teeth and other wearing surfaces of large castings have been hardened locally for many years; however, the development of suitable fixtures and equipment for heating by both gas and electrical means has greatly extended this field.

Flame Hardening. — The oxyacetylene flame is a convenient medium for rapid local heating.[46, 47] Depending on the nature of the part to be hardened, either the torch system or the work itself may be moved at such a rate that the surface temperature of the metal will be raised above the critical temperature range. Cylindrical parts are rotated before a stationary flame. An air jet or liquid spray following the torch is used to quench-harden the surface. The relatively cool metal in the interior hastens cooling at the surface in local heating processes, whereas surface cooling is always delayed when uniformly heated steel is quenched because of the conduction of heat from the hot core. The depth of flame hardening may be less than $\frac{1}{16}$ inch to about $\frac{1}{4}$ inch depending on the thickness of the section and the service requirements. Because only a small proportion of the metal is heated over the critical range, distortion caused by dimensional changes during transformation is often less than obtained by general heating and quenching methods.

In order to insure satisfactory and consistent results by the flame hardening method, the character and intensity of the flame, distance from the surface of the work, speed of travel, and timing of the quench must all be under perfect control. One inherent disadvantage of flame hardening of slowly rotating cylindrical parts is the softening or tempering effect of the flame at the final position. It is sometimes possible to heat and quench the entire surface of cylindrical parts by rapid rotation during both heating and quenching, thus eliminating the overlapping effect.

A flame hardened product is somewhat similar to a case hardened part (see Chapter X) in that it consists of a hard wear resisting surface over a tougher and more ductile core. However, case hardening by carburizing, nitriding, etc., involves a change in the chemical composition as well as the structure of the surface, while the oxyacetylene flame is generally adjusted to prevent either oxidation or carburization. It is obvious that readily hardenable steels must be used for flame hardening rather than the low carbon steels used for carburizing purposes. Both carbon and alloy steels with 0.35 to 0.70% carbon are suitable for flame hardening. Higher carbon steels are more susceptible to surface cracking. In any case the

hardening operation should be followed by a low temperature tempering or drawing treatment to relieve quenching stresses.

Among the applications of flame hardening are gear teeth, cams, roll neck bearings, wheel treads, rail ends, excavating bucket lips and teeth, wrenches, crankshafts and many other machine parts.
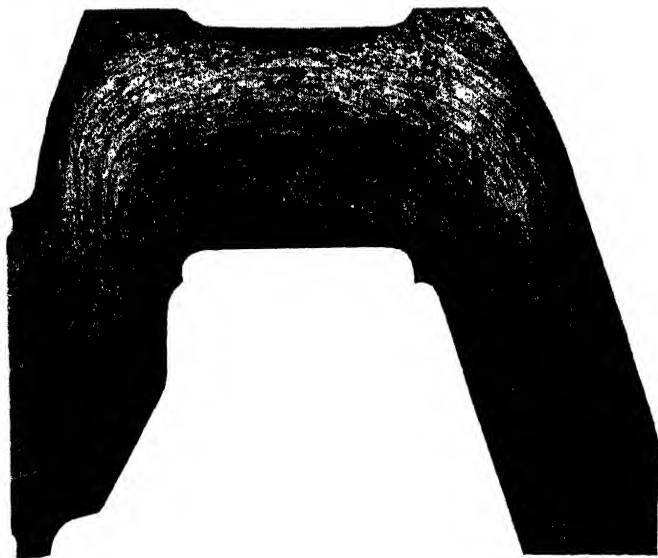
FIG. 142. — DEEPLY ETCHED SECTION OF A CRANKSHAFT SHOWING A CRANKPIN BEARING HARDENED BY THE TOCCO PROCESS. (*Courtesy of W. E. Benninghoff, Metals and Alloys.*[50])

Local Heating by Electricity. — There are three distinct methods of heating steel by electricity: first, radiation from electrical resistors in the walls of a conventional electric furnace; second, generation of heat within the stock by passage of a high current at low voltage; third, generation of heat within the stock by high frequency induction methods. All three methods are used for general heating of steel and the second and third are also used for local heating.

Direct heating by passage of current is much less expensive than heating by induction, especially from the standpoint of the equipment necessary, but both methods depend on improved quality of product, flexibility of operation, and other special considerations to compete with conventional furnace practice using electric heat, gas, or oil. Resistance heating is useful in hardening local sections of

stampings, forgings, and castings; however, its principal application is in general heating for hardening or forging of bars, tubing, and parts having uniform cross sections.[48] Because of the rapid but uniform heating throughout the section, internal stresses caused by temperature gradients are low and surface scaling is considerably reduced compared with ordinary furnace heating practice. The uniformity of temperature is especially desirable for forging operations.

Induction heating is also well adapted to heating for forging operations, especially for upsetting of bars and tubes. The ends of bars 3 inches in diameter may be heated uniformly to 2200° F. in two minutes by inserting in a special high frequency induction coil.[47] This extremely rapid heating is the result of hysteresis and eddy currents induced within the steel.

An outstanding application of induction heating of solid steel is the process for surface hardening crankshaft bearings.[49, 50] A high frequency current applied to the bearing section for a few seconds produces heating at the surface to any desired depth. The temperature rises only slightly above the Ac_3 critical temperature because of the change in the reaction of the steel to the magnetic field as the temperature rises. At the proper instant water is sprayed on the heated surface through holes in the inductor blocks surrounding the bearing. A deeply etched cross section of a crankpin bearing hardened by the patented Tocco induction heating process is shown in Fig. 142. The extent of the heated zone is so closely controlled that the fillets remain soft while the bearing surface is hardened to 60 Rockwell "C." This is a distinct advantage over general hardening methods since good ductility at the fillets tends to relieve the stress concentrations which are normally present at these locations, thereby reducing the possibility of fatigue failure without sacrificing wear resistance at the bearing surface. Another advantage over general hardening is the greatly reduced distortion during heat treatment.

As in flame hardening, a steel with good hardenability must be used. A modified S.A.E. 1050 is generally preferred where the higher core strength of alloy steels is not required. This process holds promise of wide application to innumerable cylindrical parts requiring surface hardening.

REFERENCES

1. "Glossary of Selected Iron and Steel Terms and Definitions of Terms Relating to Heat Treatment Operations," Metals Handbook, 3, Am. Soc. for Metals, Cleveland (1939)

2. Epstein, S., "The Alloys of Iron and Carbon, Vol. I — Constitution," Chapter VI, The Engineering Foundation, McGraw-Hill Book Company, New York (1936)

3. Sauveur, A., "The Metallography and Heat Treatment of Iron and Steel," McGraw-Hill Book Company, New York (1935)

4. Mehl, R. F., and Wells, C., "Constitution of High-purity Iron-carbon Alloys," Trans. Am. Inst. Mining and Met. Eng., 125, 429 (1937)

5. Davenport, E. S., and Bain, E. C., "Transformation of Austenite at Constant Subcritical Temperatures," Trans. Am. Inst. Mining and Met. Eng., 90, 117 (1930)

6. Bain, E. C., "On the Rates of Reactions in Solid Steel," Trans. Am. Inst. Mining and Met. Eng., 100, 13 (1932) — and — "Factors Affecting the Inherent Hardenability of Steel," Trans. Am. Soc. for Steel Treating, 20, 385 (1932)

7. Davenport, E. S., Roff, E. L., and Bain, E. C., "Microscopic Cracks in Hardened Steel, Their Effects and Elimination," Trans. Am. Soc. for Metals, 22, 289 (1934)

8. Vilella, J. R., Guellich, G. E., and Bain, E. C., "On Naming the Aggregate Constituents in Steel," Trans. Am. Soc. for Metals, 24, 225 (1936)

9. Zimmerman, J. G., Aborn, R. H., and Bain, E. C., "Some Effects of Small Additions of Vanadium to Eutectoid Steel," Trans. Am. Soc. for Metals, 25, 755 (1937)

10. "The Process and Result of Austenite Transformation at Constant Temperature," Metals and Alloys, 8, 22 (1937)

11. "Constitution of Steel," Metal Progress, 28, 21 (Aug. 1935) — and — 27, 24 (Feb. 1935); 33, 473 (1938); 34, 59 and 149 (1938) — see also — Metals and Alloys, 9, 223 and 230 (1938)

12. Grossmann, M. A., "Hardenability, As It Affects Heat Treated Parts," Metal Progress, 33, 373 (1938)

13. Burns, J. L., Moore, T. L., and Archer, R. S., "Quantitative Hardenability," Trans. Am. Soc. for Metals, 26, 1 (1938)

14. Mehl, R. F., "The Physics of Hardenability," Hardenability of Alloy Steels, Am. Soc. for Metals, Cleveland (1939)

15. Payson, P., and Hodapp, W., "Austempering of S.A.E. Alloy Steels Not Always Advantageous," Metal Progress, 35, 358 (1939)

16. Schaefer, A. O., "Large Steel Forgings — Notes on Modern Manufacture," Metal Progress, 35, 237 (1939)

17. Bain, E. C., and Vilella, J. R., "Austenitic Grain Size in Steel," Metals Handbook, 754, Am. Soc. for Metals, Cleveland (1939)

18. Shepherd, B. F., "The P–F Characteristic of Steel," Trans. Am. Soc. for Metals, 22, 979 (1934)

19. "Standard Grain Size Chart for Classification of Steels, A.S.T.M. Designation E19–33," Am. Soc. for Testing Mats., Standards, 761 (1936) — and — "Tentative Grain Size Chart for Classification of Steels, A.S.T.M. Designation E19–38T," Proc. Am. Soc. for Testing Mats., 38, Part I, 1296 (1938)

20. Herasymenko, P., "Austenitic Grain Size Marked by Preferential Oxidation," Metal Progress, 30, 69 (Sept. 1936)

21. Tobin, H., and Kenyon, R. L., "Austenitic Grain Size of Eutectoid Steel," Trans. Am. Soc. for Metals, **26**, 133 (1938)

22. McQuaid, H. W., "McQuaid-Ehn Test," Metals Handbook, 750, Am. Soc. for Metals, Cleveland (1939)

23. Dorn, J. E., and Harder, O. E., "Relation of Pre-treatment of Steels to Austenitic Grain Growth," Trans. Am. Soc. for Metals, **26**, 106 (1938)

24. McQuaid, H. W., "A Study of the Effect of the Aluminum Addition on the Structure of a Quenched Carbon Steel," Trans. Am. Soc. for Metals, **25**, 490 (1937)

25. Morris, M. J. R., and McQuaid, H. W., "Effect of the Silicon and Aluminum Addition," Hardenability of Alloy Steels, Am. Soc. for Metals, Cleveland (1939)

26. Brophy, G. R., and Parker, E. R., "Influence of Aluminum on the Normality of Steel," Trans. Am. Soc. for Metals, **25**, 315 (1937)

27. Chipman, J., "Aluminum as a Deoxidizer," Metals Handbook, 809, Am. Soc. for Metals, Cleveland (1939)

28. McQuaid, H. W., "The Use of Aluminum for the Control of Grain Size in Commercial Steels," Metals Handbook, 810, Am. Soc. for Metals, Cleveland (1939)

29. Bain, E. C., "Discussion of M. A. Grossmann's paper: 'On Grain Size and Grain Growth,'" Trans. Am. Soc. for Steel Treating, **21**, 1106 (1933)

30. McQuaid, H. W., "Effect of McQuaid-Ehn Grain-Size on Hardness and Toughness of Automotive Steels," Trans. Am. Soc. for Metals, **22**, 1017 (1934)

31. Jominy, W. E., and Boegehold, A. L., "A Hardenability Test for Carburizing Steel," Trans. Am. Soc. for Metals, **26**, 574 (1938)

32. Jominy, W. E., "Hardenability Tests," Hardenability of Alloy Steels, Am. Soc. for Metals, Cleveland (1939)

33. Schempp, R., "Quality Tool Steel is Tested for Hardenability," Metal Progress, **30**, 68 (Aug. 1936)

34. Grossmann, M. A., Asimow, M., and Urban, S. F., "Hardenability, Its Relation to Quenching," Hardenability of Alloy Steels, Am. Soc. for Metals, Cleveland (1939)

35. Luerssen, G. V., "Hardenability in Light Sections," Hardenability of Alloy Steels, Am. Soc. for Metals, Cleveland (1939)

36. Digges, T. G., "Effect of Carbon on the Hardenability of High Purity Iron-Carbon Alloys," Trans. Am. Soc. for Metals, **26**, 408 (1938)

37. French, H. J., and Klopsch, O. Z., "Quenching Diagrams for Carbon Steels in Relation to Some Quenching Media for Heat Treatment," Trans. Am. Soc. for Steel Treating, **6**, 251 (1924)

38. Burns, J. L., and Riegel, G. C., "Hardenability of Plain Carbon Steels," Hardenability of Alloy Steels, Am. Soc. for Metals, Cleveland (1939)

39. Bullens, D. K., and Metallurgical Staff Battelle Memorial Institute, "Steel and Its Heat Treatment," Chapter 12, John Wiley and Sons, New York (1938)

40. "Salt Baths," 317 — and — Schenck, R. B., "Lead Baths," 311, Metals Handbook, Am. Soc. for Metals, Cleveland (1939)

41. Bellis, A. E., "Molten Salt Baths Internally Heated," Metal Progress, **30**, 68 (Dec. 1936)

42. Scott, H., "Quenching Media," Metals Handbook, 328, Am. Soc. for Metals, Cleveland (1939) — and — "The Problem of Quenching Media for the Hardening of Steel," Trans. Am. Soc. for Metals, **22**, 577 (1934)

43. Zavarine, I. N., "Quenching in Water, Brine, and Oil," Metal Progress, **27**, 43 (April 1935)

44. Pilling, N. B., and Lynch, T. D., "Cooling Properties of Technical Quenching Liquids," Trans. Am. Inst. Mining and Met. Eng., **62,** 665 (1920)
45. French, H. J., "The Quenching of Steels," Am. Soc. for Steel Treating, Cleveland (1930) — and — "The Quenching and Tempering of Nickel Alloy Steels," (Section VI, Data Sheet No. 4), The International Nickel Company (1934)
46. Seemann, A. K., "Surface Hardening with the Oxy-Acetylene Flame," Metal Progress, **32,** 247 (1937)
47. Rolf, R. L., "Selective Hardening with the Oxyacetylene Flame," — and — Howat, J. T., "Flame Hardening," The Welding Journal, **17,** 6 and 23 (Feb. 1938)
48. Peters, F. P., "Direct Electric Resistance Heating of Steel," Metals and Alloys, **8,** 281 (1937)
49. Cone, E. F., "Crankshafts and Other Parts Surface Hardened by Inductive Heat," Metals and Alloys, **9,** 1 (1938) — and — **8,** 149 (1937)
50. Tran, M. A., and Benninghoff, W. E., "Differential Hardening by Induction," Trans. Am. Soc. for Metals, **25,** 935 (1937)

CHAPTER IX

LOW AND MEDIUM ALLOY STEELS

The study of alloy steels is complicated by the wide variety of compositions now in use, and by the unusual characteristics which may be imparted by large additions of alloying elements. However, the low and medium alloy steels are structurally similar to plain carbon steels and respond to heat treatment in much the same manner, therefore we may draw heavily on the principles of Chapters VII and VIII in studying this large and highly important group of alloy steels.

There is no satisfactory division point between low and medium alloy steels based entirely on chemical composition; however, steels containing less than about 1.5% of alloying elements such as Cr, Ni, Mo, V, W, Al, and Cu may be considered to be low alloy steels, as distinguished from medium alloy steels which may contain up to 5.5% total alloying elements. Considering the fact that certain elements may be ten times as effective as others in changing the characteristics of the steel, the arbitrary nature of this method of classification should be apparent. On the other hand, the high alloy steels usually contain more than 12% total alloying elements, therefore they constitute a distinct class readily distinguished from medium alloy steels. (See Chapter XIV.)

From a functional standpoint a somewhat more satisfactory classification may be made. Low alloy steels are frequently used for structural purposes in the hot rolled condition or after a simple normalizing treatment. When fabricated by forming and welding it is important that the carbon content should also be low, generally under 0.12%. Other low carbon low alloy steels are carburized and the higher carbon grades are hardened by quenching.

Medium alloy steels are generally used for machine parts, although certain high strength structural grades are also made. The low carbon medium alloy steels for machine parts are ordinarily

carburized and the higher carbon grades are almost always hardened by quenching treatments.

The high alloy steels are used for special purposes requiring exceptional corrosion, heat, or wear resistance, or other unusual properties.

Pearlitic Alloy Steels. — The low and medium alloy steels are also known as " pearlitic " steels because their normal or slowly cooled microstructures contain pearlite (plus ferrite or cementite in most cases). All of the plain carbon steels also fall in the pearlitic classification; for example, S.A.E. 1020, whose microstructure is predominately ferrite, and 1.4% carbon steel, containing excess cementite, are known as pearlitic steels.

Although qualitatively similar, medium alloy steels surpass the plain carbon steels in hardenability and in their capacity for developing high tensile properties and toughness, especially when heat treated in large sections. For a given carbon content the alloy steels as a class have lower critical cooling rates than carbon steels (they shift the " S " curve to the right), hence they may be hardened by less drastic quenching, which reduces distortion and internal stresses. On the other hand, many of the alloy steels and especially the higher alloy grades are more susceptible to cracking when drastically quenched.

There are a number of other generalities which apply to nearly all of the low and medium alloy steels. First, most alloying elements impart definite grain refining tendencies, the principal exceptions being silicon, which tends to coarsen the grain size of low carbon high silicon steels, and aluminum, which promotes coarsening when present in *large amounts.*

Second, the alloying elements lower the carbon content of the eutectoid, so that a certain alloy steel containing 0.65% carbon may be structurally equivalent to a 0.80% carbon plain carbon steel in that each would contain practically 100% pearlite in its normal structure. In fact, it will be observed later that certain alloy steels with 0.50 to 0.60% carbon may be substituted for carbon tool steels or spring steels such as S.A.E. 1090.

Third, the microstructures of most medium alloy steels, both in the slowly cooled and heat treated conditions, are so similar to those

of carbon steels that only specialists who are thoroughly familiar with the characteristics imparted by the individual alloying elements can differentiate between different types, and then only in special cases.   An expert spark tester * can sort out mixed steels with much more certainty and dispatch than a microscopist.   Both rely upon chemical analysis for accurate quantitative results.   It should be remembered, however, that chemical analysis and spark testing tell nothing of the processing or heat treating history of the alloy.

Classification of Alloying Elements by Their Effect on the Structure. — The alloying elements ordinarily used in steels may be classified in several different ways.   In this section they will be grouped according to the manner in which they occur in the microconstituents of the steel.

V   Mo   W   Cr   Ti   Cb

The elements vanadium, molybdenum, tungsten, chromium, titanium, and columbium are strong carbide formers and occur as carbides in steel provided there is sufficient carbon present to satisfy the requirements.   When the carbon content is low or the alloy content is high enough to insure an excess after carbide formation, all of these elements are capable of dissolving in the iron matrix to form solid solutions.   It has been found that upon increasing the proportion of any of these alloying elements dissolved in the iron, the alpha phase region increases in extent in the equilibrium diagram at the expense of the gamma phase region.   Relatively large additions of some of these elements result in complete disappearance of the gamma phase and stability of the alloyed alpha iron at all temperatures up to the melting point.   (See page 471.)

Al   Si   Zr   V   Ti   Cr

Of the elements listed in the first group, vanadium and titanium (and to a lesser extent chromium) have a rather definite tendency to oxidize and may appear in the steel as oxides or other nonmetallic

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* Steels give characteristic sparks at the grinding wheel depending upon their carbon and alloy contents.[1]   Differences in carbon content are fairly easy to distinguish and various alloy grades may be classified by experienced observers.

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V   Mo   W   Cr   Ti   Cb

The elements vanadium, molybdenum, tungsten, chromium, titanium, and columbium are strong carbide formers and occur as carbides in steel provided there is sufficient carbon present to satisfy the requirements. When the carbon content is low or the alloy content is high enough to insure an excess after carbide formation, all of these elements are capable of dissolving in the iron matrix to form solid solutions. It has been found that upon increasing the proportion of any of these alloying elements dissolved in the iron, the alpha phase region increases in extent in the equilibrium diagram at the expense of the gamma phase region. Relatively large additions of some of these elements result in complete disappearance of the gamma phase and stability of the alloyed alpha iron at all temperatures up to the melting point. (See page 471.)

Al   Si   Zr   V   Ti   Cr

Of the elements listed in the first group, vanadium and titanium (and to a lesser extent chromium) have a rather definite tendency to oxidize and may appear in the steel as oxides or other nonmetallic

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* Steels give characteristic sparks at the grinding wheel depending upon their carbon and alloy contents.[1] Differences in carbon content are fairly easy to distinguish and various alloy grades may be classified by experienced observers.

particles as well as in the form of carbides. To these may be added aluminum, silicon, and zirconium which are readily oxidized and are frequently observed in steel as nonmetallic inclusions such as oxides and silicates and in some cases as nitrides. All elements in the oxide forming group (Al, Si, Zr, V, Ti, Cr) are capable of dissolving in the iron matrix and, like the carbide formers, they favor the extension of the alpha phase regions of the equilibrium diagrams. Of the nine elements mentioned thus far, all tend to favor the retention of the alpha phase, several are exclusively or predominately carbide formers, others have tendencies towards both carbide and oxide formation, while aluminum, silicon, and zirconium form only oxide or other nonmetallic particles.

Ni Cu Co Mn

There remains a third group of elements consisting of nickel, copper, cobalt, and manganese. These alloys are soluble in both alpha * and gamma iron, and, with the exception of manganese, they do not, ordinarily, appear either as carbides or as oxides in steel. In contrast with the elements in the carbide and oxide forming groups, these alloys extend the gamma phase regions of the equilibrium diagrams. Manganese is rather difficult to classify in that it definitely forms carbides, oxides, and sulphides, yet it favors the formation of the gamma phase.

Phosphorus might have been placed in the oxide group, however it ordinarily occurs dissolved in the matrix rather than as an oxide. It does not form carbides and has a strong tendency to stabilize the alpha phase.

Sulphur occurs as FeS and MnS in steel and does not fit into any of the above classifications.

Characteristics Imparted to the Steel by Elements in the Three Structural Classifications. — Considering the carbide forming elements first, it will be remembered that austenite is capable of dissolving rather large amounts of iron carbide, Fe_3C, and that the solubility increases at high temperatures. The same is true with respect to carbides such as Cr_4C, Mn_3C, and VC; however, it is possible to

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* Copper is soluble in ferrite in amounts less than about 0.70%. Steels with higher copper contents are subject to precipitation of metallic copper particles.

obtain alloys with such high carbon and alloy contents that the solution of the carbides will be incomplete at ordinary heat treating temperatures.   The carbides and other compounds in high speed tool steel, for example, are so plentiful that heating to 2350° F. for quenching leaves perhaps one-third of the carbides and compounds undissolved.   Undissolved or excess carbide particles contribute directly to the hardness, wear resistance, and cutting qualities of tool steels and of some of the medium alloy machine steels; however, they reduce the hardenability of the matrix or the ease of hardening by quenching.   (To be considered in more detail later.)

Nearly all of the alloy machine steels contain special carbide particles which are dissolved in heat treatment but are precipitated in the tempering operation, thus contributing directly to the hardness, strength, and wear resistance of the product.   The advantages of alloy over iron carbides in tempered steels lie in the greater resistance to precipitation and coalescence of the special alloy carbides. This property is indicated by the fact that higher tempering temperatures are generally required to soften alloy steels than to soften plain carbon steels a similar amount.   Considering steels tempered to a given hardness value, the higher temperatures used for many alloy grades permit greater relief of quenching stresses, thus contributing to the superior toughness of these steels.   The slower coalescence of the carbides in certain alloy grades is also effective in increasing their creep strength at elevated temperatures.

Although we speak of vanadium carbide, chromium carbide, etc., it is not likely that they exist as such in the low and medium alloy steels.   It is believed that the alloy carbides are associated with iron carbide as solid solutions or complex carbides which might be written $(Fe,Cr)_3C$, for example.

Elements such as nickel and copper which remain dissolved in the matrix have a definite strengthening and toughening effect but probably do not add to the abrasive hardness of the steel.   This also applies to elements in the ferrite forming group when they are present in solid solution rather than as carbides or oxides.   The relative strengthening effects of alloying elements dissolved in ferrite are included in Table XXXV.[2]

Ordinarily the matrix containing the alloying elements in solid

solution will be ferritic since the amount of nickel or other austenite forming elements present in low and medium alloy steels will be insufficient to stabilize the austenitic structure at room temperature except under special conditions. Nevertheless, the critical temperature range of nickel and manganese medium alloy steels is lowered to

TABLE XXXV.   TRENDS OF INFLUENCE OF THE ALLOYING ELEMENTS

	As Dissolved in Ferrite: **Strength**	As Dissolved in Austenite: **Hardenability**	As Undissolved Carbide in Austenite: **Fine Grain, Toughness**	As Dispersed Carbide in Tempering: **High Temp. Strength (and Toughness)**	As Fine Dispersion of Nonmetallic: **Fine Grain, Toughness**
Al	Mild	Mild	None	None	Very Strong
Cr	Mild	Moderate	Strong	Moderate	Slight
Co	Strong	Negative	None	None	None
Cb	?	?	Strong	Strong	None
Cu	Mild	Mild	None	None	None
Mn	Strong	Moderate	Mild	Mild	Slight
Mo	Moderate	Strong	Strong	Strong	None
Ni	Mild	Mild	None	None	None
P	Strong	Mild	None	None	None
Si	Moderate	Moderate	None	None	Moderate
Ta	Moderate?	Strong?	Strong	Strong	None
Ti	?	Strong?	Very strong	Little*?	Moderate?
W	Moderate	Strong	Strong	Strong	None
V	?	Very Strong	Very strong	Very Strong	Moderate?
Zr	?	?	None?	None?	Strong?

* As a result of very slight solubility.

Courtesy of E. C. Bain, Metals Handbook.[2]

an appreciable extent as may be seen by comparing S.A.E. T1340 and 2340 with 1040 in Table XXXVIII, page 276. On the other hand, the ferrite stabilizing elements tend to raise the Ac_1 critical temperature. (Compare S.A.E. 4140, 5140, and 6140 with 1040 in Table XXXVIII.) These changes in the critical temperatures have the practical effect of altering the heat treating temperatures.

The characteristics imparted by the group of oxide forming elements have already been discussed in the preceding chapter in which it was pointed out that oxide and other nonmetallic inclusions inhibit grain growth and promote shallow hardening. It should be remembered, however, that any of these elements may be present

dissolved in the matrix and that vanadium and chromium, for example, are more likely to be present as carbides than as oxides.

Classification on the Basis of Effect on Grain Growth Tendency. — The principles of grain growth inhibition have already been mentioned. Aluminum is the outstanding example of an element whose influence on grain growth is based on finely dispersed nonmetallic particles. It is a relatively cheap addition but great care must be exercised in its use to insure satisfactory mechanical properties. It has been highly successful as an addition to wrought steels but is less generally used in steel foundries.

Vanadium and titanium are highly effective in controlling grain size because of the stability of their carbides at elevated temperatures. Vanadium is by far the most expensive addition but is highly regarded because of its beneficial effect on impact toughness and other mechanical properties.

The mechanism of grain refinement by zirconium is not yet fully revealed but there is no uncertainty in the results obtained.

All other alloying elements contribute more or less to grain refinement. Probably the least effective are those completely soluble in ferrite and austenite; however, the increased sluggishness of alloyed austenite may be effective in reducing grain growth as well as in retarding transformation.

Classification on the Basis of the Effect on Hardenability. — A second general classification may be made on the basis of the effect of the alloying elements on hardenability as measured by changes in the critical cooling rate. Here the task becomes quite difficult because of the dual effects of certain elements, and the complex behaviors which result when more than one alloying element is used.

It is now generally believed that nearly all of the useful alloying elements delay the transformation of austenite on cooling (shift the " S " curve to the right) and thereby promote deep hardening provided they are dissolved in the austenite at the quenching temperature. According to Bain (third column of Table XXXV) cobalt is the only important alloying element known to have a negative effect on hardenability when dissolved in austenite at the quenching temperature.

As undissolved carbides (fourth column), or as finely dispersed

nonmetallic particles (last column), several of the elements inhibit grain growth and in this way reduce hardenability. Obviously the net effects of these elements on hardenability will vary with the metallurgical history and final heat treatment of a given steel. As an example, the effect on hardenability of the strong carbide forming element, vanadium, is complicated by the fact that ordinary hardening temperatures may not be sufficiently high to dissolve the stable carbides present.[3] This reduces the carbon content as well as the alloy content of the austenite, thereby reducing the tendency towards deep hardening. Furthermore, the undissolved carbides present act as nuclei for austenite transformation which also tends to increase the critical cooling rate. It has been shown that higher quenching temperatures and longer time at temperature increase hardenability in such cases.

The primary effect of metallic aluminum in solid solution is believed to be improved hardenability, yet the very strong grain refining influence of aluminum oxide particles far outweighs the hardening action when only small amounts of aluminum are present so that ordinarily aluminum treated steels tend towards shallow hardening.

It is interesting to note that nickel, copper, and cobalt, elements which extend the range of the gamma phase on the equilibrium diagrams, have relatively small effects in stabilizing the gamma phase (delaying the transformation of austenite) upon quenching, whereas the elements vanadium, molybdenum, and tungsten, which extend the alpha region under equilibrium conditions, have a strong tendency to delay austenite transformation and increase hardenability upon quenching.

The above classification and discussion is necessarily based on the individual effects of the several elements considered, assuming that carbon is present in normal amounts. When combinations of alloying elements are used, somewhat different results may be expected. It is well known, for example, that certain elements intensify the action of others in the complex alloy steels.

Classification of Alloying Elements by Cost. — A third general classification of the alloying elements may be made on the basis of unit cost. The approximate cost per pound of metal introduced into

the finished steel by means of special alloying additions is given in Table XXXVI. Since several of the alloys may be introduced in different forms, representative addition agents were selected in preparing this table. In general, the ferroalloys contain large concentrations of the special alloying elements with iron as the base, carbon and other elements being present as impurities. A number of complex additions containing more than one special element are used but are not included in the table.

These figures are only approximate, the principal variable aside from fluctuation in market price being the recovery of metal in the steel making process. This will depend upon the type of furnace, furnace conditions, and the time of making the additions. In general, the recovery is greater in the electric furnace than in basic open hearth practice. The estimated recoveries were taken mainly from " Metals Handbook "[4] and the prices of raw materials from " American Metal Market " for Jan. 4, 1939.

According to this classification Mn, Si, Cu, Cr, and P are the least expensive of the commonly used alloying elements. Molybdenum is also a moderately priced addition since only small amounts, usually less than 0.35%, are needed to produce the desired results in medium alloy steels. Although even smaller amounts of vanadium are ordinarily used, less than 0.20%, vanadium is an expensive addition.

The low carbon ferrochromium is used in making stainless steels. Less costly ferrochromium additions with higher carbon contents are suitable for low and medium alloy steels. Small additions of columbium and titanium are used to stabilize stainless steels for certain purposes. (See Chapter XIV.)

Aluminum is used principally for grain size control and deoxidation purposes and very little metallic aluminum remains in the steel. The principal exceptions are nitriding steels which contain about 1% aluminum.

Titanium and zirconium additions are also made for removing oxygen and nitrogen and for inhibiting grain growth.

The real costs of the alloying elements depend to a large extent upon the relative amounts required; consequently, the base prices for the S.A.E. steels (see page 278) give a better comparison of

TABLE XXXVI.   RELATIVE COSTS OF ALLOY ADDITIONS

FA = Ferroalloy corresponding to the element
B.O.H. = Basic open hearth practice
B.E. = Basic electric practice

Element	Addition Agent	Assumed Point of Addition	Assumed Recovery	Cost in $/lb. of Alloying Element in Finished Steel
V	FA, 50% V	Ladle, B.O.H.	90%	3.00
Cb	FA, 50% Cb	Ladle, B.E.	90%	2.50
Ti	FA, 25% Ti Low carbon [a]	Ladle, B.E.	50%	2.46
W	FA, 80% W	Furnace	High	1.75
Co	Co, 98%	Furnace	High	1.36
Mo	Ca Molybdate, 40% Mo	Furnace, B.O.H.	95%	0.84
Ni	Ni, 99%	Furnace	Very high	0.36
P	FA, 24% P	Ladle, B.O.H.	75%	0.21
Cr	FA, 70% Cr, under 0.1% C	Furnace, B.E.	95%	0.21
Cr	FA, 70% Cr, 4–6% C	Ladle, B.O.H.	90%	0.12
Cu	Cu, 100%	Furnace	Very high	0.10
Si	FA, 50% Si	Ladle, B.O.H.	80% [b]	0.087
Mn	FA, 80% Mn	Ladle, B.O.H.	80% [b]	0.076
Al	Al, 99%	——	[c]	[c]
Ti	FA, 17% Ti, 7.0% C	——	[d]	[d]
Zr	FA, 35% Zr, 50% Si	——	[e]	[e]
(Fe)	——	——	——	(0.021)

[a] For addition to stainless steels. (See Chapter XIV.) A cheaper grade is used for deoxidation in B.O.H. practice.

[b] In quantities added for alloying purposes. Recovery may be very much lower when used for deoxidation.

[c] Used mainly for deoxidation, $0.20 per lb.

[d] Used as a deoxidizer and scavenger, $0.43 per lb. of Ti.

[e] Used as a deoxidizer and scavenger, $0.38 per lb. of Zr.

the costs of various alloy steels than can readily be obtained from Table XXXVI.

The Society for Automotive Engineers Classification of Alloy Steels. — The chemical compositions of the well known S.A.E. steels are reproduced in Table XXVI, page 188, and Table XXXVII fol-

TABLE XXXVII. CHEMICAL COMPOSITIONS OF S.A.E. STEELS

MANGANESE STEELS *

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.
T1330	0.25–0.35	1.60–1.90	0.040	0.050
T1335	0.30–0.40	1.60–1.90	0.040	0.050
T1340	0.35–0.45	1.60–1.90	0.040	0.050
T1345	0.40–0.50	1.60–1.90	0.040	0.050
T1350	0.45–0.55	1.60–1.90	0.040	0.050

NICKEL STEELS *

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Nickel Range
2015	0.10–0.20	0.30–0.60	0.040	0.050	0.40–0.60
2115	0.10–0.20	0.30–0.60	0.040	0.050	1.25–1.75
2315	0.10–0.20	0.30–0.60	0.040	0.050	3.25–3.75
2320	0.15–0.25	0.30–0.60	0.040	0.050	3.25–3.75
2330	0.25–0.35	0.50–0.80	0.040	0.050	3.25–3.75
2335	0.30–0.40	0.50–0.80	0.040	0.050	3.25–3.75
2340	0.35–0.45	0.60–0.90	0.040	0.050	3.25–3.75
2345	0.40–0.50	0.60–0.90	0.040	0.050	3.25–3.75
2350	0.45–0.55	0.60–0.90	0.040	0.050	3.25–3.75
2515	0.10–0.20	0.30–0.60	0.040	0.050	4.75–5.25

NICKEL-CHROMIUM STEELS *

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Nickel Range	Chromium Range
3115	0.10–0.20	0.30–0.60	0.040	0.050	1.00–1.50	0.45–0.75
3120	0.15–0.25	0.30–0.60	0.040	0.050	1.00–1.50	0.45–0.75
3125	0.20–0.30	0.50–0.80	0.040	0.050	1.00–1.50	0.45–0.75
3130	0.25–0.35	0.50–0.80	0.040	0.050	1.00–1.50	0.45–0.75
3135	0.30–0.40	0.50–0.80	0.040	0.050	1.00–1.50	0.45–0.75
3140	0.35–0.45	0.60–0.90	0.040	0.050	1.00–1.50	0.45–0.75
X3140	0.35–0.45	0.60–0.90	0.040	0.050	1.00–1.50	0.60–0.90
3145	0.40–0.50	0.60–0.90	0.040	0.050	1.00–1.50	0.45–0.75
3150	0.45–0.55	0.60–0.90	0.040	0.050	1.00–1.50	0.45–0.75
3215	0.10–0.20	0.30–0.60	0.040	0.050	1.50–2.00	0.90–1.25
3220	0.15–0.25	0.30–0.60	0.040	0.050	1.50–2.00	0.90–1.25
3230	0.25–0.35	0.30–0.60	0.040	0.050	1.50–2.00	0.90–1.25

* Silicon range of all S.A.E. basic open hearth alloy steels shall be 0.15–0.30. For electric and acid open hearth alloy steels, the silicon content shall be 0.15 minimum.

TABLE XXXVII.   (*Continued*)

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Nickel Range	Chromium Range
3240	0.35–0.45	0.30–0.60	0.040	0.050	1.50–2.00	0.90–1.25
3245	0.40–0.50	0.30–0.60	0.040	0.050	1.50–2.00	0.90–1.25
3250	0.45–0.55	0.30–0.60	0.040	0.050	1.50–2.00	0.90–1.25
3312	max. 0.17	0.30–0.60	0.040	0.050	3.25–3.75	1.25–1.75
3325	0.20–0.30	0.30–0.60	0.040	0.050	3.25–3.75	1.25–1.75
3335	0.30–0.40	0.30–0.60	0.040	0.050	3.25–3.75	1.25–1.75
3340	0.35–0.45	0.30–0.60	0.040	0.050	3.25–3.75	1.25–1.75
3415	0.10–0.20	0.30–0.60	0.040	0.050	2.75–3.25	0.60–0.95
3435	0.30–0.40	0.30–0.60	0.040	0.050	2.75–3.25	0.60–0.95
3450	0.45–0.55	0.30–0.60	0.040	0.050	2.75–3.25	0.60–0.95

MOLYBDENUM STEELS *

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Chromium Range	Nickel Range	Molybdenum Range
4130	0.25–0.35	0.50–0.80	0.040	0.050	0.50–0.80	—	0.15–0.25
X4130	0.25–0.35	0.40–0.60	0.040	0.050	0.80–1.10	—	0.15–0.25
4135	0.30–0.40	0.60–0.90	0.040	0.050	0.80–1.10	—	0.15–0.25
4140	0.35–0.45	0.60–0.90	0.040	0.050	0.80–1.10	—	0.15–0.25
4150	0.45–0.55	0.60–0.90	0.040	0.050	0.80–1.10	—	0.15–0.25
4320	0.15–0.25	0.40–0.70	0.040	0.050	0.30–0.60	1.65–2.00	0.20–0.30
4340	0.35–0.45	0.50–0.80	0.040	0.050	0.50–0.80	1.50–2.00	0.30–0.40
X4340	0.35–0.45	0.50–0.80	0.040	0.050	0.60–0.90	1.50–2.00	0.20–0.30
4615	0.10–0.20	0.40–0.70	0.040	0.050	—	1.65–2.00	0.20–0.30
4620	0.15–0.25	0.40–0.70	0.040	0.050	—	1.65–2.00	0.20–0.30
4640	0.35–0.45	0.50–0.80	0.040	0.050	—	1.65–2.00	0.20–0.30
4815	0.10–0.20	0.40–0.60	0.040	0.050	—	3.25–3.75	0.20–0.30
4820	0.15–0.25	0.40–0.60	0.040	0.050	—	3.25–3.75	0.20–0.30

CHROMIUM STEELS *

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Chromium Range
5120	0.15–0.25	0.30–0.60	0.040	0.050	0.60–0.90
5140	0.35–0.45	0.60–0.90	0.040	0.050	0.80–1.10
5150	0.45–0.55	0.60–0.90	0.040	0.050	0.80–1.10
52100	0.95–1.10	0.20–0.50	0.030	0.035	1.20–1.50

* Silicon range of all S.A.E. basic open hearth alloy steels shall be 0.15–0.30. For electric and acid open hearth alloy steels, the silicon content shall be 0.15 minimum.

TABLE XXXVII.   (*Continued*)

CHROMIUM-VANADIUM STEELS *

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Chromium Range	Vanadium Min.	Desired
6115	0.10–0.20	0.30–0.60	0.040	0.050	0.80–1.10	0.15	0.18
6120	0.15–0.25	0.30–0.60	0.040	0.050	0.80–1.10	0.15	0.18
6125	0.20–0.30	0.60–0.90	0.040	0.050	0.80–1.10	0.15	0.18
6130	0.25–0.35	0.60–0.90	0.040	0.050	0.80–1.10	0.15	0.18
6135	0.30–0.40	0.60–0.90	0.040	0.050	0.80–1.10	0.15	0.18
6140	0.35–0.45	0.60–0.90	0.040	0.050	0.80–1.10	0.15	0.18
6145	0.40–0.50	0.60–0.90	0.040	0.050	0.80–1.10	0.15	0.18
6150	0.45–0.55	0.60–0.90	0.040	0.050	0.80–1.10	0.15	0.18
6195	0.90–1.05	0.20–0.45	0.030	0.035	0.80–1.10	0.15	0.18

TUNGSTEN STEELS *

S.A.E. No.	Carbon Range	Manganese, Max.	Phosphorus, Max.	Sulphur, Max.	Chromium Range	Tungsten Range
71360	0.50–0.70	0.30	0.035	0.040	3.00–4.00	12.00–15.00
71660	0.50–0.70	0.30	0.035	0.040	3.00–4.00	15.00–18.00
7260	0.50–0.70	0.30	0.035	0.040	0.50–1.00	1.50– 2.00

SILICON-MANGANESE STEELS

S.A.E. No.	Carbon Range	Manganese Range	Phosphorus, Max.	Sulphur, Max.	Silicon Range
9255	0.50–0.60	0.60–0.90	0.040	0.050	1.80–2.20
9260	0.55–0.65	0.60–0.90	0.040	0.050	1.80–2.20

CORROSION AND HEAT RESISTING ALLOYS

S.A.E. No.	Carbon, Max.	Manganese, Max.	Silicon, Max.	Phosphorus, Max.	Sulphur, Max.	Chromium Range	Nickel Range
30905	0.08	0.20–0.70	0.75	0.030	0.030	17.00–20.00	8.00–10.00
30915	0.09–0.20	0.20–0.70	0.75	0.030	0.030	17.00–20.00	8.00–10.00
51210	0.12	0.60	0.50	0.030	0.030	11.50–13.00	—
X51410	0.12	0.60	0.50	0.030	0.15–0.50	13.00–15.00	—
51335	0.25–0.40	0.60	0.50	0.030	0.030	12.00–14.00	—
51510	0.12	0.60	0.50	0.030	0.030	14.00–16.00	—
51710	0.12	0.60	0.50	0.030	0.030	16.00–18.00	—

* Silicon range of all S.A.E. basic open hearth alloy steels shall be 0.15–0.30. For electric and acid open hearth alloy steels, the silicon content shall be 0.15 minimum.

Reprinted from the S.A.E. Handbook, 1938 edition, issued by the Society of Automotive Engineers, Inc., 29 West 39th St., New York City.

TABLE XXXVIII.   APPROXIMATE CRITICAL TEMPERATURES FOR
S.A.E. STEELS

S.A.E. Number	Ac_1 °F	Ac_3 °F	S.A.E. Number	Ac_1 °F	Ac_3 °F	S.A.E. Number	Ac_1 °F	Ac_3 °F
Carbon Steels			Nickel Steels			Molybdenum Steels		
1010	1350	1605	2015	1375	1575	4130	1395	1485
1015	1355	1585	2115	1345	1525	X4130	1395	1480
1020	1355	1570	2315	1300	1440	4135	1395	1475
1025	1355	1545	2320	1285	1420	4140	1380	1460
1030	1350	1495	2330	1275	1400	4150	1365	1395
1035	1345	1475	2335	1275	1375	4340	1350	1425
1040	1340	1455	2340	1280	1360	4345	1345	1415
X1040	1340	1450	2345	1280	1350	4615	1335	1485
1045	1340	1450	2350	1280	1340	4620	1335	1470
X1045	1335	1420	2515	1250	1420	4640	1320	1430
1050	1340	1425	2520 *	1240	1390	4650 *	1315	1410
X1050	1335	1400				4815	1300	1440
1055	1340	1425				4820	1300	1440
X1055	1335	1400						
1060	1340	1410	Nickel-Chromium Steels			Chromium Steels		
1065	1340	1385	3115	1355	1500	5120	1410	1540
X1065	1335	1380	3120	1350	1480	5140	1370	1440
1070	1345	1370	3125	1350	1465	5150	1330	1420
1075	1350	1365	3130	1345	1460	52100	1340	1415
1080	—	1360	3135	1340	1445			
1090	—	1360	3140	1355	1415	Chromium-Vanadium Steels		
1095	—	1360	X3140	1350	1430			
10150 *	—	1355	3145	1355	1395	6115	1420	1550
			3150	1335	1380	6120	1410	1545
			3215	1350	1465	6125	1400	1490
Free Cutting Steels			3220	1350	1460	6130	1390	1485
1112	1355	1590	3230	1340	1435	6135	1390	1480
1120	1355	1550	3240	1335	1425	6140	1390	1455
X1315	1345	1520	3245	1345	1400	6145	1390	1450
			3250	1340	1375	6150	1385	1450
			3312	1330	1435	6195	1370	1425
Manganese Steels			3325	1335	1400	Tungsten Steels		
T1330	1325	1480	3330 *	1320	1380	7260	1360	1430
T1335	1315	1460	3335	1310	1360			
T1340	1315	1435	3340	1290	1380	Silicon-Manganese Steels		
T1345	1315	1410	3415	1330	1425			
T1350	1310	1400	3435	1290	1380	9255	1400	1500
T1360 *	1305	1405	3450	1290	1360	9260	1400	1500

TABLE XXXVIII.   (*Continued*)

In most instances the critical temperatures were found by examining the microstructures of quenched discs 1 inch in diameter by $\frac{3}{8}$ inch thick, taken from the furnace in 20° F. increments.   The grain sizes of the steels are unknown.   The rate of heating was about the same as in commercial annealing treatments, thus the critical points are for practical application and do not, necessarily, coincide with the equilibrium transformation temperature.   The Ar$_1$ and Ar$_3$ critical temperatures are also given in the original.

* This is not in the official S.A.E. list.

Courtesy of M. J. R. Morris, R. Sergeson, and G. W. Gable.
Condensed from " Metal Progress," Vol. 32, p. 462 (1937).

lowing.   This classification is widely used in its original form and has been adopted, with modifications, by many producers and users of steels.   Although the classification applies mainly to the so-called machine steels, a limited number of tool and high alloy steels are included.

Specific recommendations for carburizing and heat treating these steels will be found in the " S.A.E. Handbook."   With few exceptions the higher carbon S.A.E. steels should be quenched in oil while S.A.E. Nos. T1330, 2330, 2335, 3125, 3130, 4130, X4130, 52100, 6125, 6130, and 6195 may be quenched in water when the size and shape of the piece permits.   Irregular and small sections should be quenched in oil to prevent cracking.

Mechanical Properties of S.A.E. Steels. — The mechanical properties of heat treated S.A.E. steels are often represented in graphical form as in Figs. 143 and 144 from the " S.A.E. Handbook." The following notation from the Handbook indicates their purpose and limitations: " The following physical property charts should not be considered in any way a part of the standard S.A.E. steel specifications or as giving absolute values.   They are intended only as a guide to proper heat treatment of the steels and to indicate conservative physical properties that may be expected of standard test specimens 0.505 x 2 in., machined from rolled bars up to 1½ in. diameter or square."

Although a large proportion of the heat treated parts made from these steels have relatively light sections and fall within the size limit given above, it should be remembered that steels heat treated

BASE COSTS OF REPRESENTATIVE IRON AND STEEL PRODUCTS
(See page 477 for stainless steels)

Structural shapes	2.10¢/lb.
Plates	2.10 ”
Standard rails, ton	$40.00
Common iron bars	2.23¢/lb.
Single refined genuine wrought iron	4.08 ”
Double refined genuine wrought iron	5.63 ”
Soft steel bars	2.25 ”
Rail steel bars	2.10 ”
Reinforcing bars	2.05 ”
Cold finished bars, carbon	2.70 ”
Cold finished bars, alloy	3.40 ”
Sheets, hot rolled	2.15¢/lb.
Sheets, cold rolled	3.20 ”
Sheets, 24 gage galvanized	3.50 ”
Enameling sheets, 20 gage	3.35 ”
Tin mill black, 29 gage	3.05 ”
Tin plate, per base box, 100 lb.	$5.00
Annealed fence wire	2.95¢/lb.
Galvanized fence wire	3.35 ”
Wire nails	2.45 ”
Basic pig iron	$20.50/gross ton
Bessemer pig iron	21.50 ” ”
No. 2 foundry pig iron	21.00 ” ”
Malleable pig iron	21.00 ” ”
Heavy melting steel scrap	$15.75/gross ton

ALLOY STEEL BARS

Hot rolled, base, open hearth	2.80¢/lb.
Cold finished, base, open hearth	3.40 ”
Hot rolled, base, electric furnace	3.30 ”

S.A.E. Steel	Alloy Differential	S.A.E. Steel	Alloy Differential
10xx	0.00¢/lb.	32xx	1.35¢/lb.
T13xx	0.10 ”	33xx	3.80 ”
20xx	0.35 ”	34xx	3.20 ”
21xx	0.75 ”	41xx	0.55 ”
23xx	1.55 ”	4340	1.85 ”
25xx	2.25 ”	X4340	1.65 ”
31xx	0.70 ”	46xx	1.10 ”

S.A.E. Steel	Alloy Differential	S.A.E. Steel	Alloy Differential
48xx	2.00¢/lb.	92xx (spring rounds) .	0.40¢/lb.
51xx (0.6–0.9% Cr) ..	0.35 "	92xx (spring flats)	0.15 "
51xx (0.8–1.1% Cr) ..	0.45 "	C–Mo* (0.10–0.20% Mo)	0.35 "
51xx (spring flats)	0.15 "	C–Mo* (0.15–0.25% Mo)	0.40 "
61xx (bars)	1.20 "	C–Mo* (spring flats) ...	0.15 "
61xx (spring flats) ..	0.85 "		

* Not in the official S.A.E. list.

Taken mainly from "American Metal Market" for Jan. 4, 1939. The base price of steel bars has varied between 1.50 and 2.45¢/lb. in the past ten years.

in heavier sections develop lower strength and hardness as shown in Figs. 145 and 146 from the "S.A.E. Handbook."

In general, the alloy additions reduce the critical cooling rate, which tends to counteract the effect of increased mass on reducing the actual cooling rate. The average strength and hardness of a large section are thus increased and the difference between core and surface properties is reduced. The effect of any specific alloy addition depends on its ability to reduce the critical cooling rate or to shift the "S" curve to the right. (See column 3, Table XXXV, page 268.)

The S.A.E. mechanical property charts have been available for a number of years as a conservative guide to the mechanical properties of machine steels. Informed engineers and metallurgists using these charts appreciate the fact that the steels represented are subject to variations in properties obtained upon heat treatment. The rather wide permissible ranges of carbon and alloy content are a primary contributing factor. Austenitic grain size at the heat treating temperature is another factor which may vary widely depending on the deoxidation process used in making the steel and the use of elements such as aluminum, titanium, and zirconium (which are not covered by the composition specifications) for grain size control. In a given case, should the tensile strength of an S.A.E. steel actually coincide with the values given on the chart for a certain treatment, it is likely that the elongation and reduction of area will be appreciably higher than shown, or for a given elongation or reduction of area the strength properties will be higher than shown. For these reasons one may prefer to use the data found in handbooks and cata-
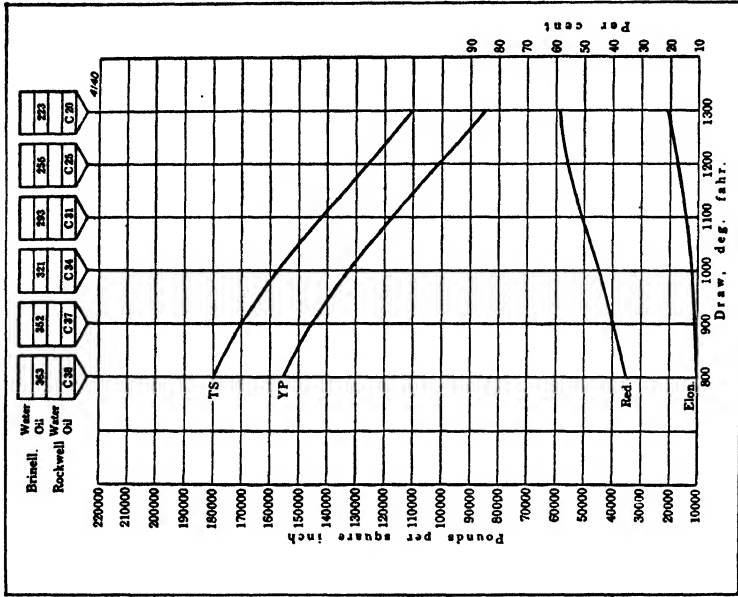
FIG. 144. — PROPERTIES OF S.A.E. 4140, NORMALIZED AT 1650°–1750° F, QUENCHED IN OIL FROM A TEMPERATURE OF 1525°–1625° F, DRAWN AT THE INDICATED TEMPERATURES. (Courtesy of the Society of Automotive Engineers. S.A.E. Handbook, 1938.)
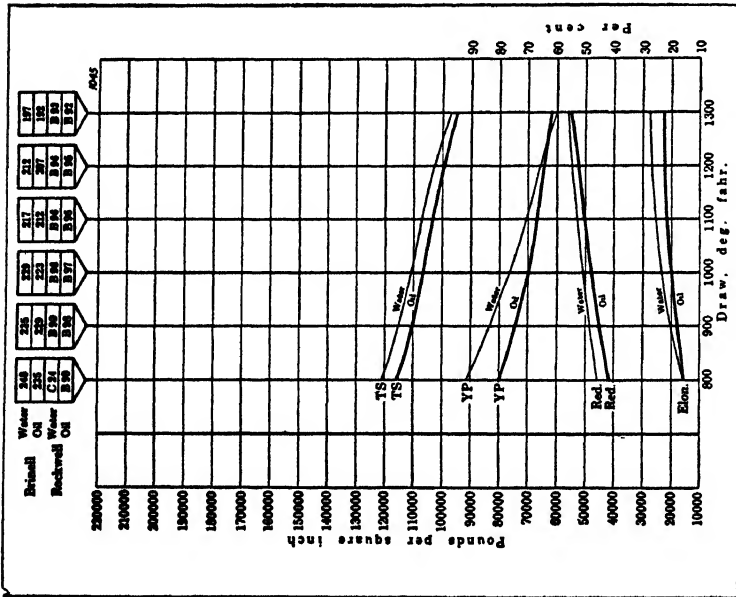
FIG. 143. — PROPERTIES OF S.A.E. 1045, QUENCHED IN WATER OR OIL FROM A TEMPERATURE OF 1475°–1525° F. AND DRAWN AT THE INDICATED TEMPERATURES. (Courtesy of the Society of Automotive Engineers. S.A.E. Handbook, 1938.)
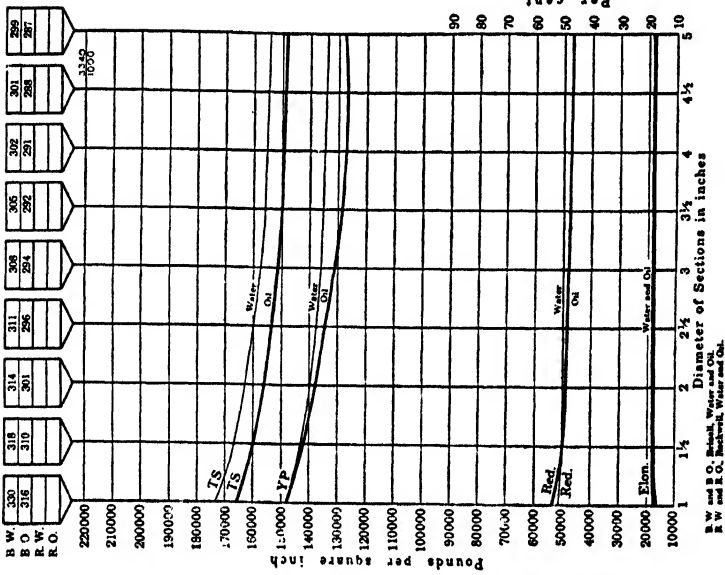
FIG. 146. — PROPERTIES OF S.A.E. 3340 HEAT TREATED IN VARIOUS SECTION SIZES BY NORMALIZING AT 1600°–1700° F., QUENCHING IN WATER OR OIL FROM 1450° F. FOLLOWED BY DRAWING AT 1000° F. (*Courtesy of the Society of Automotive Engineers. S.A.E. Handbook, 1938.*)
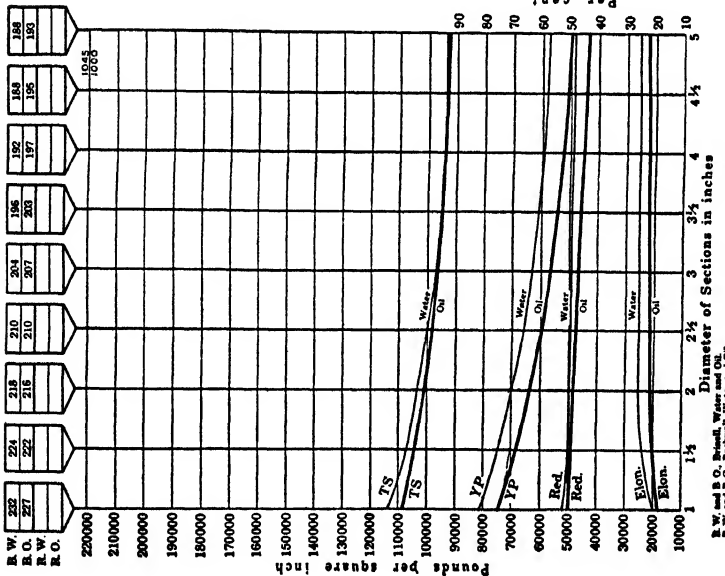
FIG. 145. — PROPERTIES OF S.A.E. 1045 HEAT TREATED IN VARIOUS SECTION SIZES BY QUENCHING IN WATER OR OIL FROM 1500° F. FOLLOWED BY DRAWING AT 1000° F. (*Courtesy of the Society of Automotive Engineers. S.A.E. Handbook, 1938.*)

logs issued by the steel manufacturers or in " Steel Physical Properties Atlas." [5]

The close relationship between the mechanical properties of the various S.A.E. steels when heat treated to the same strength has recently been emphasized by Janitzky and Baeyertz [6] and Boegehold.[7] The data plotted in Fig. 147 are for the same steels listed in Fig. 148. (The individual points are identified in " Metals Handbook.")

The higher strength obtained at a given tempering temperature, or the greater resistance to tempering of certain alloy steels, is clearly indicated in Fig. 148. The steels containing molybdenum, such as S.A.E. 4340 and 4145, are outstanding in this respect.

Janitzky and Baeyertz conclude that the S.A.E. machine steels are substantially equivalent on the basis of mechanical properties as determined by the tensile test under the following conditions:

"(1) If the sections are so chosen that the test pieces of each steel are *hardened throughout on quenching.*

"(2) If the test pieces are *tempered to the same tensile strength,* irrespective of the tempering temperature necessary to produce that tensile strength.

"(3) If a tensile strength of 200,000 psi. *is not exceeded.* Under the above conditions of quenching, the tempering temperature required to obtain a given tensile strength depends both on the alloy and the carbon content of the steel and must be determined experimentally."

The strength-hardness relationship is also included in the above survey. The points fall on a straight line and for practical purposes the following factors may be used:

Brinell × 480 = tensile strength at 100,000 lb./sq. in. tens. str.
    "   × 490 =   "   "   " 200,000 " / " "   "   "
    "   × 500 =   "   "   " 300,000 " / " "   "   "

Boegehold's results, Fig. 149, show somewhat more deviation but fit the same average line.

In order to obtain the higher hardness and tensile values it is necessary to temper the carbon steels at much lower temperatures
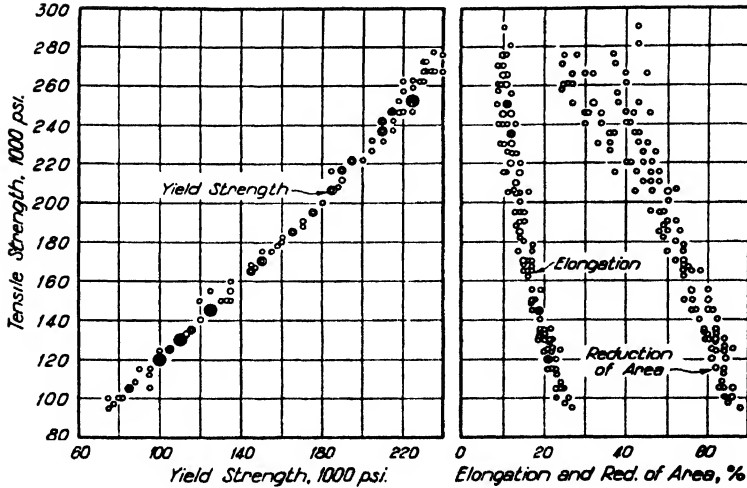
FIG. 147. — RELATIONSHIPS BETWEEN MECHANICAL PROPERTIES OF WATER AND OIL HARDENING S.A.E. STEELS. BARS 1 INCH IN DIAMETER WERE NORMALIZED, QUENCHED IN WATER OR OIL, ACCORDING TO S.A.E. RECOMMENDATIONS, THEN TEMPERED AT VARIOUS TEMPERATURES FROM 400°–1300° F. (*Courtesy of E. J. Janitzky and M. Baeyertz, Metals Handbook.*[6])
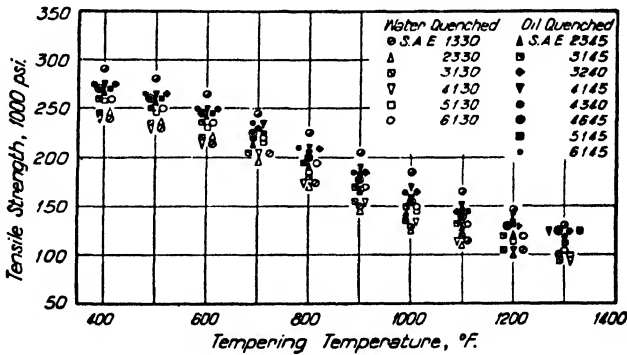
FIG. 148. — CHANGE IN TENSILE STRENGTH WITH INCREASE IN TEMPERING TEMPERATURE FOR WATER AND OIL HARDENING STEELS. THE SPECIMENS WERE HELD AT TEMPERATURE FOR 30 MINUTES. (*Courtesy of E. J. Janitzky and M. Baeyertz, Metals Handbook.*[6])

than the alloy steels.   This results in lower ductility as shown in Fig. 150.   (Note also the spread in reduction of area at high strengths in Fig. 147.)   Alloy steels are required to develop high ductility at high hardness and strength, an important factor to be considered in
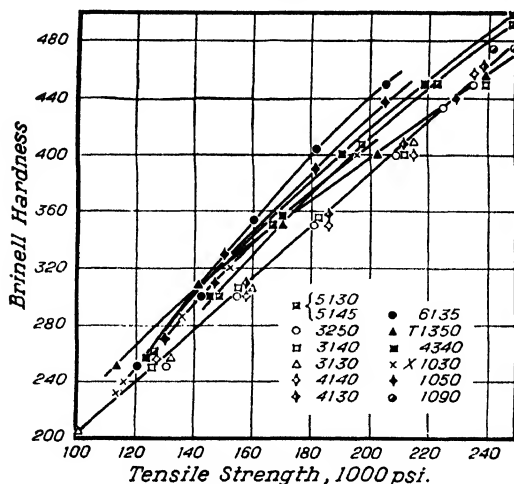
FIG. 149. — TENSILE STRENGTH VERSUS BRINELL HARDNESS OF CARBON AND ALLOY S.A.E. STEELS HEAT TREATED IN SMALL SECTIONS. (*Courtesy of A. L. Boegehold, Metal Progress.*[7])
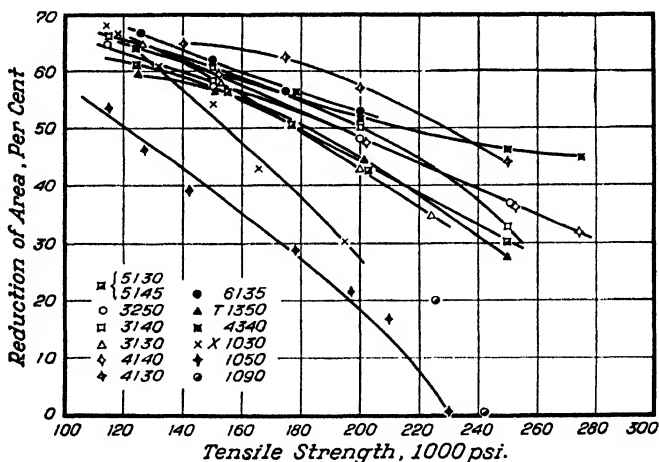
FIG. 150. — TENSILE STRENGTH VERSUS REDUCTION OF AREA OF CARBON AND ALLOY S.A.E. STEELS HEAT TREATED IN SMALL SECTIONS.

(ALL BARS HEAT TREATED AS 1 INCH ROUNDS EXCEPT THOSE OF X1030, 1050, AND 1090, WHICH WERE 0.510 INCH SPECIMENS, AND 4130, WHICH WERE 1.33 INCH ROUNDS. ALL SPECIMENS OIL QUENCHED EXCEPT S.A.E. 4130 AND X1030, WHICH WERE WATER QUENCHED.)

(*Courtesy of A. L. Boegehold, Metal Progress.*[7])

the case of highly stressed machine parts of irregular contour in which stress concentrations cannot be avoided. Good ductility is believed to be essential for redistribution of stresses in such parts just as in structural designs using mild steels.

It might be assumed that the carbon steels would be entirely satisfactory for parts subjected to high alternating or repeated stresses provided they are working well below their endurance limit. Since the endurance limit of smooth specimens is approximately equal to one-half the tensile strength, high endurance limits can be obtained even with carbon steels by using low drawing temperatures. Here again, however, the relatively low ductility and high internal stress of carbon steels treated to very high strengths militates against relief of concentrated stresses, hence the effective endurance limit may be materially lower than that determined for smoothly finished specimens, and fatigue failures are more likely to occur in high strength carbon steels than in suitable alloy steels drawn to the same hardness and strength.

Another limitation of carbon steels for high strength heat treated parts is their inferior machinability. When machining operations must be carried out on heat treated parts over 300 Brinell, several of the alloy steels are preferred over carbon steels.

It should be remembered that the preceding arguments in favor of alloy steels apply principally to parts that must be heat treated in large sections and to highly stressed parts, especially those having sharp fillets, screw threads, holes, projections, or other stress raisers.

Merit Indices. — When comparing the mechanical properties of steels as an aid in selecting an analysis and heat treatment suitable for a particular application, it is natural to attempt to combine the data for each steel in a single numerical index. Some of the merit indices which have been proposed are [8, 9]:

(1) Tensile Strength $\times$ Elongation

(2) Elastic Limit $\times$ Reduction of Area

(3) Elastic Limit $\times$ $\sqrt[3]{\text{Elongation} \times \text{Reduction of Area}}$

(4) Tensile Strength $\times$ Izod Impact Value

(5) $\dfrac{\frac{1}{2}\,(\text{Tensile Strength} + \text{Elastic Limit})\,(\text{Elongation})}{(100 - \text{Reduction of Area})}$

These formulas have, for the most part, been judged to be inadequate and misleading as a basis for selection of a suitable composition of steel.

Boegehold [9] in considering merit indices as a means of appraising steels for automotive uses made the following statements:

" A steel which has already been tested and proved satisfactory in some car part may be gaged by some merit index, like those just mentioned, and as long as succeeding shipments of the same specification of steel meet the merit index requirement, it can reasonably be expected to respond normally to processing treatments and to function satisfactorily in service.

" It does not follow, however, that some other steel having just as high a merit index will be suitable for the same part, or that a steel with a higher index will have a longer life in that part. The reason for this is that the stresses in a tensile test specimen are not similar to the stresses in a single important part of an automobile. Furthermore, the merit index obtained by means of a tensile test is expressed in values that are a function of a type of failure which does not occur when automobile parts fail (excepting when the car hits a telegraph pole). The tensile test specimen does not break until it has elongated and become smaller in diameter at one point. Broken car parts, on the other hand, almost invariably show no elongation or reduction of area.

" One variety of this kind of fracture is fatigue failure, caused by a large number of stress applications above the endurance limit but below the elastic limit. Being below the elastic limit there is no apparent flow of metal. Another variety of fracture results from a breaking stress applied at a notch, and fracture through a notch or groove will also occur without any manifestation of ductility."

After considering the possibility of using merit indices for appraising S.A.E. steels for tractor rear axle shafts, Knowlton [10] concluded that none of the mathematical formulas were suitable for determining either the best composition or tempering temperature for a given application. In the following statement Knowlton questions in particular the emphasis placed on plasticity (as expressed by elongation and reduction of area) in most of the indices used:

". . . it is the writer's thought that the useful function of plas-

ticity is to permit sufficient deformation of the more highly stressed fibres, in a nonuniformly stressed article, to bring about a more favorable distribution of stress. If this is the case there is a useful limit of plasticity depending upon the design of the article and the type of service which it must withstand. A higher degree of plasticity would not increase the merit of the steel so far as that particular application is concerned." [10]

It is generally conceded that the surest method for determining the most suitable composition and treatment for a given part is comparison under actual service conditions. Where this is impractical the next best method is comparison under test conditions closely approximating service conditions; for example, torsion impact and torsion fatigue tests on full size axles are among those considered useful in studying axle steels. In most cases it is much more economical to make tests using standard specimens and testing procedures, thus introducing the problem of interpreting the results in terms of a particular application. In order to accomplish this some standard or standards of merit must be established, and whether or not mathematical relations are used depends upon the observer.

Carbon Ranges in Alloy Steels. — Before considering the S.A.E. steels by types, it will be well to recall the significance of the various carbon ranges, as discussed on page 196. The carbon contents of the pearlitic alloy steels are, in general, somewhat lower than those of plain carbon steels intended for similar applications. The upper carbon limit for safety in water quenching is also lower than for plain carbon steels. (The medium carbon S.A.E. alloy steels which are recommended for water quenching were listed on page 276. Oil quenching is recommended even for these steels when the sections are small or irregular.) The alloy steels containing 0.35 to 0.55% carbon are almost always quenched in oil.

Manganese as an Alloying Element. — Manganese is now recognized as one of the most useful alloying elements. In addition to its marked effect on the hardenability of steels upon quenching, it also improves the hot working characteristics and the strength and wear resistance of hot rolled, annealed, or normalized steels. (See page 171.)

The carbide forming tendencies of manganese are mild compared with chromium and molybdenum, for example, hence the principal effect of manganese is strengthening of the matrix in which it is dissolved. When present in austenite it renders its transformation to pearlite or martensite sluggish, delaying the Ar_1 transformation to pearlite and lowering the temperature of martensite formation at Ar''.[11] The resulting effect on hardenability is generally considered to be strong compared with silicon, nickel, copper, and other elements not in the carbide forming group; however, it will be noted that Bain describes its effect on hardenability as " moderate " in Table XXXV on page 268.

Manganese steels have a tendency towards " temper brittleness " when cooled slowly after tempering. This property is described further on page 299.

Manganese has little if any tendency to inhibit grain growth of austenite, consequently grain controlling additions of aluminum, titanium, or other elements are needed to insure fine grain size if high heat treating temperatures are used.

The action of manganese as a cleansing agent, and the formation of MnS and Mn_3C have already been mentioned. Manganese differs from the other important carbide forming elements such as chromium and vanadium in that it tends to lower the critical temperature range, and at high concentrations it extends the austenitic phase region.[12] It will be seen later that in these respects it resembles nickel.

Nearly all steels contain manganese, ranging from a few hundredths of one per cent in certain low carbon sheet steels to about 2%, in the pearlitic steels, and up to 14% in austenitic manganese steels. (See Chapter XI, page 380.)

The use of manganese as the principal alloying element in steels to be heat treated (S.A.E. T13xx series) is a comparatively recent development. In fact, steel users were at one time prejudiced against using manganese because of the known brittleness of early manganese steels. While manganese has never been considered comparable to nickel as a toughener, the brittleness often attributed to this element has been proved to be caused by using too high a combined carbon and manganese content. Various combinations of

these elements which have been used in successful steels are outlined in Fig. 151.[13] Although not indicated, high strength steel castings and heat treated forging steels also occupy the region between 1% and 2% manganese and less than about 0.5% carbon. The S.A.E. 13xx steels are also in this range. Their use for carburized
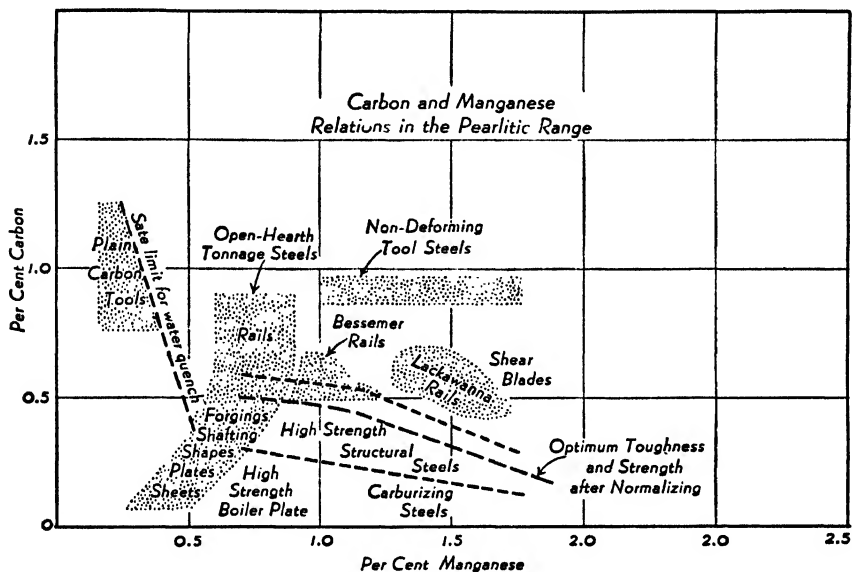
Fig. 151. — CARBON AND MANGANESE RELATIONS IN THE PEARLITIC RANGE. (*Courtesy of E. E. Thum, Proc. Am. Soc. for Testing Mats.*[13] MODIFIED BY THE ADDITION OF THE NON-DEFORMING TOOL STEEL RANGE AND OMISSION OF THE MARTENSITIC AND AUSTENITIC REGIONS.)

and other heat treated parts has increased greatly in recent years, largely because their superior mechanical properties and machinability are obtained with only slight increase in cost over plain carbon steels. As in the case of other medium alloy types, the manganese steels must be heat treated in order to develop fully their potential mechanical properties. While ordinary quenching and tempering methods give best results where applicable, normalizing treatments are also used to obtain desirable properties in rails, large forgings, and steel castings.[14, 15]

In order to prevent cracking in heat treatment, the manganese contents of water quenching tool steels must be carefully limited, as

indicated in Fig. 151.   However, an important type of oil hardening non-deforming tool steel contains over 1% manganese.

Although the Bessemer rails indicated in Fig. 151 are no longer produced, a considerable tonnage of rail and wheel steel contains approximately 0.75% carbon and 0.70% manganese, and certain special rails contain up to 2.0% manganese with correspondingly lower carbon contents.

Silicon as an Alloying Element. — It is difficult to define the lower limit of silicon content as an alloying element.   The semi-killed and rimming steels used in large quantities for sheet, plate, wire, pipe, and structural shapes contain little or no silicon.   Nearly all killed steels are partially deoxidized by ferrosilicon, which leaves residual silicon contents of 0.15 to 0.50%.   The silicon content of steel castings is generally near the upper limit.   The division point between silicon treated steels and silicon alloy steels is indefinite.

Silicon strengthens ferrite and increases the hardenability when dissolved in austenite.   It raises the critical temperatures and has a moderate effect in retarding grain growth in carbon steels.   Although silicon improves resistance to scaling at elevated temperatures, it greatly increases the tendency towards decarburization of the surface.   Silicon killed steel ingots tend to pipe deeply and require special attention to obtain good surface on rolled products.

Low cost silicon steels of somewhat greater strength than plain carbon steels have been used for constructional purposes for many years.   The grade most widely used in this country is a low silicon type with 0.20–0.40% C, 0.60–0.90% Mn, and 0.20–0.35% Si. The German " Freund Steel " contains about 0.12% C, 0.40–0.70% Mn, and 0.75–1.50% Si.[16]   Various intermediate silicon contents are used in other grades.   (See " Alloys of Iron and Silicon." [17])

An alloy known as silicon-manganese spring steel, S.A.E. 9255–9260, has been widely used for automobile leaf springs.   The manganese content is important for promoting hardenability, and should be on the high side for deep hardening.   Compared with carbon spring steels heat treated to the same tensile strength, approximately 200,000 lb. per sq. in., the silicon-manganese steel has a higher yield strength.   This property, higher elastic ratio, is characteristic of the alloy steels when heat treated to high strengths.

S.A.E. 9255 permits greater latitude in heat treating temperatures than S.A.E. 1090 but coarsens more readily than S.A.E. 6150, a high grade chromium-vanadium spring steel. Furthermore, silicon-manganese springs must be amply protected against decarburization, or ground after hardening to meet exacting endurance requirements. The toughness of S.A.E. 9255 is no better than that of plain carbon spring steels and less than that of S.A.E. 6150; however, the latter is much more expensive. (See page 279 for comparative base costs.)

Silicon Electrical Sheet Steels. — Steels containing up to 5% silicon and very low carbon and impurity contents have magnetic properties particularly suited to electrical applications in transformers, motors, and generators. In alternating current machines the iron circuits are very rapidly magnetized, demagnetized, magnetized with opposite polarity, and demagnetized, usually at the rate of 60 complete cycles per second. The losses in electrical energy in the magnetic circuit, called iron losses, are greater than in direct current applications. Factors which tend to reduce iron losses are:

(1) The presence of silicon in solid solution in the ferrite. (Additions of copper, aluminum, and phosphorus have also proved useful in certain alloys.)

(2) The reduction of carbon and other elements to a minimum. (Carbon, oxygen, and sulphur, elements which form compounds with iron, are particularly injurious.)

(3) Coarse grain size.

(4) High electrical resistivity. (Increased by elements which dissolve in ferrite.)

(5) Freedom from internal stress. (Produced by adequate annealing.)

These factors are interrelated; for example, silicon contributes directly to items 3 and 4. Coarse grain size is effective in reducing magnetic hysteresis, but probably increases the production of eddy currents within the grains, causing energy loss due to internal electrical heating. This effect is greatly reduced by using very thin sheets to build up laminated transformer cores and motor fields. High resistivity also reduces eddy currents.

The silicon contents used range from 0.25% for small motors for intermittent operation to nearly 5% in special transformer grades. The base costs range from 3.20¢ per lb. for " Field grade " to 7.65¢ per lb. for " Transformer extra special " (" American Metal Market " for Jan. 4, 1939).   The energy losses (in watts per lb. of iron under certain test conditions) range from approximately 2 watts per lb. for " Field grade " in 24 gage sheets, to about one-half this amount in a moderately high silicon steel of the same gage, to about one-quarter this amount for the highest quality 29 gage electrical sheets.   As the silicon increases the alloys become quite hard and brittle; for example, a fully annealed transformer sheet steel containing 4.3% silicon, 0.02% carbon, and less than 0.20% other elements had a Vickers-Brinell hardness of 230, a tensile strength of 70,000 lb. per sq. in., and an elongation of 4% in 2 inches.   However, the mechanical properties of most electrical steels are of little practical importance except from the standpoint of punching characteristics.

FIG. 152. — TRANSFORMER SHEET STEEL CONTAINING 4.3% SILICON. (Etched with 3% nital.)   SECTIONS OF TWO SHEETS ARE SHOWN AT X75.

The microstructure of the above mentioned steel is shown in Fig. 152.   The extremely coarse but homogeneous ferrite grain structure is characteristic of the high silicon grades.

Nickel Alloy Steels. — Nickel steels were the first alloy steels to be used in appreciable quantities in large engineering structures. Armor plate, highly stressed bridge members, and shafting were among the early applications.   It was found that nickel increased the strength, hardness, and toughness, without sacrificing the ductility.

The effect of nickel on the structure of pearlitic steels is compara-

tively simple in that it is completely soluble in both alpha and gamma iron, and while the amount present is sufficient to lower the critical transformation temperatures appreciably, and thereby lower the permissible heat treating temperatures (see Table XXXVIII, page 276), it is not sufficient to stabilize the austenite as in the case of certain high alloy steels to be discussed later. The critical cooling rates are reduced by nickel, making it possible to harden many sections in oil which would otherwise require water quenching.

Nickel steels and the more complex S.A.E. steels containing nickel along with other alloying elements retain good impact toughness at very low temperatures such as encountered, for example, in processing equipment for dewaxing crude oils. Carbon steels, wrought iron, and many other ferrous products are often found deficient in this respect.[18]

As in the case of the medium manganese steels, a good combination of properties may be obtained in the rolled state or by simple annealing or normalizing treatments, making it possible to improve the mechanical properties of boiler plates, structural shapes, and large forgings or castings which cannot be readily quenched. Steels of this type containing less than 0.25% carbon and from 1.5 to 3% nickel are used extensively for locomotive boilers, bolts, stay bolts, railway axles, and large forgings.

The following data by Wickenden[19] from " Metals Handbook " illustrate the effect of nickel on annealed steels.

Carbon %	Nickel %	Tensile Strength lb./sq. in.	Yield Point lb./sq. in.	Elongation in 2 in., %	Reduction of Area, %	Brinell Hardness
0.30	0.0	70 000	40 000	26	50	145
0.30	3.5	90 000	60 000	27	55	175
0.30	5.0	95 000	65 000	28	56	185

The most efficient use of the alloy addition is obtained upon quenching and drawing. The S.A.E. 23xx steels are well known for their excellent mechanical properties while the more costly S.A.E. 25xx steels, containing 5% nickel, develop still greater strength and toughness and are used for certain parts subjected to high static and impact stresses in service.

Chromium Alloy Steels. — The straight chromium steels, S.A.E. 51xx, have been used extensively for machine parts because their high hardenability, strength, and very good wear resistance are obtainable at relatively low cost.   Other low cost alloy steels of somewhat greater toughness are now becoming more popular; however, it will be noted later that a large proportion of these steels contain chromium as one of the principal elements.

The straight chromium steels are used in many carbon contents in addition to those covered by the S.A.E. specifications.   The high hardness and abrasiveness of these steels is present to a marked degree in the high carbon grade, S.A.E. 52100, which is the principal composition used for balls and bearings in assembled " frictionless " bearings.   (Carburized steels are also used, especially for roller bearings.)

The pearlitic chromium steels generally have less than 1.5 % chromium, and 1 % is the usual amount found in alloy machine steels. High chromium contents lower the critical cooling rate to such an extent that air hardening results, which introduces the possibility of cracking upon cooling after hot rolling, forging, welding, or heat treating.   Special precautions are necessary in performing these operations on the special grades of air hardening steels with chromium contents up to 3.5 % ; however, these steels are readily hardened in heavy sections and the 3 % chromium type develops high strength with good toughness and ductility upon normalizing.   Service tests on an experimental 3 % chromium rail steel indicate that satisfactory toughness as well as high wear resistance may be obtained with carbon contents of about 0.25 % rather than 0.70 % as in ordinary rail steels, or 0.60 % as in medium manganese rails.[20]

A steel containing 4 to 6 % chromium is used for oil refinery equipment operating at approximately 1000° F.   Seamless tubes, valves, fittings, etc., made from this steel give several times the life of carbon steel and cost much less than the high alloy stainless varieties.   (See Chapter XIV.)

The entire field of straight chromium steels up to 10 % chromium, as well as many of the complex alloy steels containing chromium, are treated comprehensively by Kinsel and Crafts in " Alloys of Iron and Chromium, Vol. I — Low-Chromium Alloys." [21]

Chromium-Manganese Alloy Steels. — Among the numerous complex steels containing chromium as one of the alloying elements is the comparatively new structural steel known as Cromansil containing 0.5% Cr, 1.2% Mn, and 0.8% Si.[22]   The carbon content of this steel is varied depending upon the application; however, it is generally under 0.25%.   The chromium content is believed to intensify the strengthening effect of manganese and silicon and to give better ductility than medium manganese steels of the same strength.

According to Kinsel [23], a steel containing 0.8% chromium and 0.8% manganese with a medium carbon content would be low in ductility in the rolled or normalized condition, but when either the chromium or the manganese is raised somewhat, with a corresponding lowering of the other element, so-called " balanced " compositions are obtainable which have desirable properties.   Furthermore, a composition which may have suitable properties in the normalized state may not be desirable for products which are to be quenched and drawn.   To further complicate the situation, the addition of a third or fourth alloying element may again change the balance of the major constituents.   For example, a high tensile structural steel known as Chromador which was developed in England contains approximately 0.2% C, 0.9% Cr, 0.8% Mn, and 0.3% Cu.   In the hot rolled state it develops a tensile strength of 90,000 lb. per sq. in. with 17% elongation in 2 in.[24]

A chromium-manganese steel modified by vanadium has been developed for normalized forgings requiring somewhat less strength than obtainable in hardened and drawn alloy steels.   This steel, known as " Normalloy," contains 0.25–0.45% C, 0.30–0.40% Cr, 1.05–1.30% Mn, and 0.08–0.11% V.[25, 26]   The vanadium addition refines the grain size in the rolled and normalized conditions, thus increasing the toughness.   The chromium content tends to equalize the properties throughout large sections.   This steel has been used for railroad forgings and for various automotive parts such as crankshafts, connecting rods, axle shafts, and steering arms. Properties of a normalized crankshaft containing 0.40 to 0.45% carbon have been reported as: yield point, 70,000 lb. per sq. in.; tensile strength, 110,000 lb. per sq. in.; elongation in 2 inches, 20%; reduction in area, 53%; and Brinell hardness 221.   This composi-

tion is representative of certain moderate cost steels which are being developed to give good mechanical properties after simple heat treatments.   Similar steels without the chromium content are well known as manganese-vanadium forging steels.

Alloy Structural Steels. — The S.A.E. alloy steels are generally used for machine parts.   Alloy steels for structural purposes must ordinarily be lower in cost and must be suitable for use without heat treatment, or after a normalizing treatment.   Several important structural grades including silicon, medium manganese, and chromium-manganese-silicon steels have already been mentioned.   The more costly nickel steels containing up to about $3\frac{1}{2}\%$ nickel are also used to a limited extent for certain highly stressed structural members.

A special group of low alloy structural steels has been developed for application in the transportation industries and for other purposes where light weight is a basic requirement.[27, 28, 29]   Weight reduction in railway freight and passenger cars, and in trucks, buses, and other vehicles may be effected using three different classes of materials; the light aluminum and magnesium alloys, low alloy steels, and high strength stainless steels.   Low alloy tonnage steels are by far the cheapest per pound.   Although their strength-weight ratios do not equal those of the other materials mentioned, they have been widely used for many purposes such as freight car and tank car construction.   Aluminum alloys and high strength stainless steels are factors to be considered in the construction of railway passenger cars, buses, and other vehicles; nevertheless, over 1200 rail units for passenger service have already been made of low alloy high strength steels.[30]   Where weight is of prime importance, as in metal aircraft construction, the low alloy steels have not competed, so far, with aluminum alloys and high strength stainless steels.

In general, the low alloy steels have yield points ranging from 45–65,000 lb. per sq. in. compared with 30–40,000 lb. per sq. in for mild steels. Their tensile strengths are correspondingly high, up to about 90,000 lb. per sq. in., their impact toughness is generally superior to that of corresponding carbon steels, and their ductility is only slightly lower.

In order to make the use of thinner sections practical for many constructional purposes, it was necessary to develop alloys having

corrosion resistance superior to that of ordinary mild steel. Small additions of copper have been used for this purpose for many years; however, only moderate improvement is possible by this means. Recognition of the beneficial effect of phosphorus, especially in combination with copper, and the possibility of further improvement through the use of small additions of other elements such as silicon, nickel, chromium, and molybdenum led to the de-

TABLE XXXIX.   TYPICAL LOW ALLOY HIGH STRENGTH STEELS

	Mn	Si	Cu	Ni	Mo	P*	Cr
Cor-Ten....	0.10–0.50	0.50–1.00	0.30 0.50	——	——	0.10–0.20	0.50–1.50
Man-Ten....	1.25–1.70	0.30 max.	0.20 min.	— —	——	0.04 max.	——
Sil-Ten......	0.60 min.	0.20 min.	0.20**	——	——	0.04 max.	——
Yoloy.......	0.30–0.90	0.10–0.25	0.85–1.10	1 50–2.00	——	——	——
RDS–1....	0.50 1.00	— —	0.50–1.50	0.50–1.00	0.10 min.	0.10 max.	——
RDS–1A ...	0.50 1.00	——	0.50 1.50	0.50 1.00	0.10 min.	0.04 max.	——
Hi-Steel....	0.50 0.70	0.30 max.	0.90–1.25	0.45–0.65	——	0.10–0.15	——
H.T.–50....	0.20 min.	0.10 max.	0.35 min.	0.50 min.	0.05 min.	0.05–0.15	——
A. W. Dyn-El.	0.50–0.80	——	0.30–0.50	——	——	0.06–0.10	0.25 max.
Jal-Ten	1.25–1.75	0.30 max.	0.40 min.	——	——	0.04 max.	——
Konik.	——	— —	0.10–0.30	0.30–0.50	——	——	0.07–0.30
Mayari R ..	0.50–1.00	0.05–0.50	0.50–0.70	0.25–0.75	——	0.04–0.12	0.20–1.00

* P and S range is 0.035–0.055 unless otherwise stated.
** Optional.

Courtesy of Edwin F. Cone, " Metals and Alloys." [20] (Mechanical properties and other data are given in " Metals and Alloys.")

velopment of several new structural steels. The nominal compositions of several trade-named products are listed in Table XXXIX. While all of the compositions listed are high strength alloys, the atmospheric corrosion resistance varies; for example, Man-Ten has approximately the same corrosion resistance as that of mild steel with the same copper content, while Cor-Ten is said to have from four to six times the atmospheric corrosion resistance of mild steel.

Carbon content is not given in Table XXXIX because some manufacturers vary it with the application. These steels do not depend upon carbon for their increased strength, and in general they contain less than 0.12% carbon. Higher carbon contents are undesirable where cold forming operations are necessary in fabrication. Low carbon contents also reduce air hardening after welding, which is a very important factor considering the extensive use of welding

in fabricating light weight equipment. Carbon contents up to about 0.35% may be used in some cases where unusually high strength, rather than weldability or ductility, is the principal requirement.

The alloy contents of these steels are also subject to change depending on the application. Further changes in the compositions and the introduction of new alloys are to be expected considering the relatively short time that steels of this type have been available.

It should be remembered that these are tonnage steels, supplied as sheet, plate, and structural shapes, therefore they are becoming a very appreciable part of the total steel production.

Copper Steels. — Copper contents under 0.20% are used in tonnage steels for improved resistance to atmospheric corrosion. Larger amounts up to 1.50% are found in the high yield point low alloy steels. Steels with copper contents over 0.70% offer great promise as precipitation hardening alloys. Moderately fast cooling from ordinary normalizing temperatures is sufficient to retain copper in solid solution, even in heavy sections. Subsequent precipitation in the tempering temperature range (maximum effect between 900° and 1100° F.) causes precipitation of particles of copper and increase in strength up to about 20,000 lb. per sq. in. without serious decrease in ductility. Copper contents between 1.0 and 1.5% and carbon contents under 0.55% are considered best. The principal applications thus far have been in steel castings.[31, 32] (See also Chapter XI.)

The Pearlitic Nickel-Chromium Alloy Steels. — The nickel-chromium steels had the advantage of a very early entry into the alloy steel market. Certain Cuban iron ores have the proper alloy content to yield a pig iron (Mayari iron) from which properly proportioned nickel-chromium steels are made without major alloy additions. Although nearly all nickel-chromium steels are now made by the addition of alloying elements in the usual way, these " natural " alloys influenced their early development.

The combined effect of nickel and chromium in pearlitic steels gives an excellent combination of hardenability, strength, wear resistance, and toughness at a cost lower than that of straight nickel steels of equal strength. The various nickel-chromium combinations are capable of developing tensile properties covering the en-

tire range available with alloy steels. The higher alloy types, espe-
cially S.A.E. 33xx, and 43xx (a nickel-chromium-molybdenum steel),
have excellent depth hardening characteristics in heavy sections.
(See Fig. 146.)   The high impact toughness of nickel and nickel-
chromium steels, and especially those modified by the addition of
molybdenum, is illustrated by Thum [33] using Izod Toughness-Tensile
Strength diagrams.

Temper Brittleness. — Many alloy steels are susceptible to tem-
per brittleness or low impact toughness after cooling slowly from
the tempering temperature.   In the case of steels to be used at nor-
mal or slightly elevated temperatures, susceptibility to temper brit-
tleness is not ordinarily a serious problem since rapid cooling in the
tempering operation develops the normal toughness of the part;
however, steels which are heat treated in intricate or non-uniform
sections may develop seriously high internal stresses if quenched
from the tempering heat.

Steels which are repeatedly heated and cooled in service may de-
velop room temperature brittleness, therefore compositions which
are not susceptible to temper brittleness are highly desirable for
applications involving heating to temperatures up to about 1200° F.

The mechanism by which impact toughness is affected by cooling
rate has not been satisfactorily explained.   The most acceptable
theory is probably that based on aging or dispersion hardening by
carbide particles which separate upon slow cooling but are retained
in solution upon quenching.   However, it is difficult to explain, on
this basis, the complete recovery of toughness often observed at
slightly elevated testing temperatures.

The chromium-nickel, chromium, and manganese alloy steels are
among those most likely to be susceptible to temper brittleness.
There is much evidence indicating that molybdenum in steel reduces
this tendency, the most striking of which is the effect of molybdenum
on 5% chromium steel used in the oil refining industry.   (See page
488.   The literature on temper brittleness is reviewed in reference
26 of Chapter XIV.)

Chromium-Molybdenum Alloy Steels. — Small additions of
molybdenum are used to improve plain carbon and medium man-
ganese steels, the carbon-molybdenum type, known as " Amola "

steel, being used rather extensively for automotive parts; however, the principal application of molybdenum has been in chromium, nickel, and nickel-chromium steels. The chromium-molybdenum grades, which were developed during and after the World War, are now established as strong competitors of the older nickel-chromium compositions. One of the outstanding features of chromium-molybdenum and nickel-chromium-molybdenum steels is retention of high ductility when heat treated to high strengths. (See Fig. 150.) These favorable static tensile properties are accompanied by high toughness and fatigue strength, good forming characteristics, and excellent weldability, a combination of properties which accounts for the extensive use of S.A.E. X4130 tubes, bars, and forgings in aircraft fuselage, wings, engine mounts, and landing gears.[34, 35]

Molybdenum is believed to form complex carbide particles containing iron, molybdenum, and carbon. These carbides are more stable than pure Fe_3C, consequently molybdenum steels require higher heat treating temperatures or longer holding time at temperature, compared with carbon steels, in order to dissolve the carbides and utilize the full effect of the alloy content in developing hardness and strength upon quenching. In the tempering operation higher temperatures are required to produce softening as shown in Fig. 148.

The success of carbon-molybdenum and other molybdenum steels for elevated temperature service is also associated with the stability of the carbides present. In addition to molybdenum, other carbide forming alloying elements, principally vanadium and chromium, may be used to improve the creep resistance [31, 36, 37] of steels for boilers and other structures requiring strength at elevated temperatures. When operating temperatures are such as to produce rapid oxidation of carbon and low alloy steels, the high alloy heat resisting grades are generally used. These steels retain their strength at higher temperatures than the pearlitic carbon and alloy steels. (See Chapter XIV.)

The molybdenum steels are known to be machinable at higher hardnesses than many other alloy grades, making it possible to perform certain final machining operations after heat treatment.

Many automobile parts including drive shafts, steering arms, and

high strength bolts have been made of S.A.E. 4130 and X4130. The higher carbon grades are also used for axles, drive shafts, and oil hardening gears. A summary of typical automotive uses of alloy steels will be found on page 305.

Chromium-Vanadium Alloy Steels. — The S.A.E. 61xx steels are recognized as being among the toughest and most reliable of the alloy steels. Chromium-vanadium steels are unsurpassed for leaf and coil springs, and were it not for their high cost they could be used for practically all applications of hardened and tempered springs.

Vanadium is a strong carbide forming element and it resembles molybdenum in its effect on the properties of steel. It is even more potent than molybdenum in that 0.10 to 0.20% vanadium is usually sufficient in the pearlitic steels, while somewhat higher amounts of molybdenum are used. The highly dispersed carbides and possibly oxides of vanadium are believed to be responsible for its effectiveness in obstructing grain growth.

Vanadium is quite different from molybdenum in one respect in that it is very readily oxidized in the steel making furnace or in the ladle, thus contributing to the deoxidation and cleanliness of the finished steel, which is believed to be partly responsible for the good dynamic properties obtained. Vanadium is far too expensive to be used as the principal deoxidizer and is usually added in the ladle or shortly before tapping in order to prevent excessive loss.[38]

Vanadium carbides are even more difficult to dissolve in the austenite than molybdenum carbides, therefore heat treating temperatures well over 100° F. above Ac_3 are often used for best hardenability. The vanadium steels are ordinarily very fine grained and resist coarsening at these high heat treating temperatures. However, upon heating over the coarsening temperature range very marked grain growth is likely to occur, as in the case of aluminum treated steels.

The S.A.E. 61xx steels are used for a wide variety of highly stressed machine parts ranging from low carbon carburized articles to gears and springs containing 0.50% carbon. They are also used for certain processing equipments operating at elevated temperatures up to 1000° F. Their structural stability and strength at mod-

erately high temperatures is comparable to that of the molybdenum steels and superior to most other pearlitic steels.

Vanadium steels are used for large forgings and castings because of the uniform and fine grain size obtainable after casting, forging, or simple normalizing treatments.

Other Alloy Steels. — Only well known grades of pearlitic alloy steels (also silicon electrical steel) have been mentioned in this chapter. A comprehensive treatment of complex alloy steels will be found in " Steel and Its Heat Treatment " — Vol. II.[31] Further references to alloying elements will be found in each of the remaining chapters.

Processing Schedule for an Alloy Steel Forging. — The following processing sequence illustrates some of the operations required in making machine parts from alloy steel forgings.

(1) S.A.E. 4140 (5–7 A.S.T.M. grain size) bar stock is heated to 2050° F. and forged. The blanks are to be used for making certain lever shafts for an automobile steering mechanism.

(2) Heat to 1650° F.; hold at heat for one hour; cool slowly to 800° F.; air cool to room temperature. Remove scale.

(3) Rough machine and cut threads at one end.

(4) Heat to 1550° F. with the furnace; hold at heat for one-half hour; quench in oil.

(5) Temper at 1000° F. in a circulating air type furnace.

(6) Test 10% of the parts for Brinell hardness. The hardness limits are 321 to 415 Brinell.

(7) Draw each end in a lead bath at 1250°–1275° F. for 15 minutes.

(8) Test the end of each part for Brinell hardness. The hardness limits at the ends are 175–200 Brinell.

(9) Straighten to 0.01 inch.

(10) Mill serrations (for load transmission and locking purposes) adjacent to the threaded end of the shaft. Machine the other end which will carry a hardened steel cam follower mounted in roller bearings.

(11) Finish grind.

Naturally every design calls for individual attention in establishing a suitable processing schedule, and the example used is not intended to be generally applicable.

The central portion of this lever operates in a bronze bushing and must, therefore, be reasonably hard and wear resistant (approximately 360 Brinell). The second tempering of the end sections improves their machinability and toughness. Rapid transmission of heat in the lead bath permits tempering of the ends without softening of the central bearing surface. This forging is used in a large steering assembly for a truck. The treatment for a similar but smaller lever which is carburized before hardening is given on page 331.

Carbon versus Alloy Steels. — Many of the advantages of alloy over plain carbon heat treated steels were illustrated in the foregoing pages and may be summarized as follows:

(1) Alloy steels have lower critical cooling rates than carbon steels, which improves their hardenability, especially in heavy sections.

(2) Since alloy steels are hardenable at lower cooling rates, the tendency to form surface and internal cracks is reduced, and there is generally less distortion upon quenching.

(3) Alloy steels may be tempered at higher temperatures to attain a given strength and hardness. This tends to reduce residual internal stresses, one of the factors responsible for the improved toughness and ductility of alloy over plain carbon steels.

(4) Wear resisting surfaces of gears and many carburized parts made of alloy steels may be hardened fully by oil quenching, whereas carbon steels may require water quenching to eliminate soft spots.

(5) At a given tensile strength, and especially at very high strengths, the alloy steels are tougher and more ductile and have higher yield strengths.

(6) The greater ductility at high strength reduces the danger of stress concentrations at screw threads, machining marks, etc., thereby raising the practical endurance stress limit and making the alloy steels more reliable for irregular sections.

(7) Many types of alloy steels may be machined at higher hard-

nesses after heat treatment than is possible with plain carbon steels. In some cases where alloy steels may be machined after heat treatment, carbon steels having the necessary mechanical properties must be machined before hardening and then be straightened and ground for dimensional accuracy.

(8) The danger of excessive grain growth in heat treatment is reduced in most alloy steels. It will be shown that this is especially important in carburizing.

(9) In certain grades of pearlitic alloy steels the strength is retained at moderately high operating temperatures to a greater extent than in carbon steels.

The list of advantages of carbon over alloy steels is necessarily short. However, low cost is not the only factor which accounts for the extensive use of carbon steels in machine parts. For example, carbon steel springs with 0.55 to 0.95% carbon are very successful for small to moderate size sections of simple shape without regions of high stress concentration. They are readily hardened and tempered to around 400 Brinell, and retain sufficient ductility to operate safely even at high stresses.[7] Other simple and symmetrically shaped parts may often be drastically quenched to obtain a very high surface hardness and wear resistance without the use of alloying elements.

The successful and economical operation of many plain carbon steel parts is associated with the fact that their modulus of elasticity, and consequently their stiffness, is equal to that of the best alloy steels. Furthermore, when parts are designed for high rigidity the sections are often so large or of such a shape that the working stresses are reduced to well within the safe range for carbon steels tempered to a high ductility. In such parts the shallow hardening characteristics of the carbon steels may not be a handicap if working stresses are low and only the surfaces need be hardened for wear resistance.

In many cases the importance of machinability outweighs all other requirements including strength, ductility, and toughness, hence the free-cutting grades of carbon steels are preferred over all of the alloy steels.

Steels for Automobiles. — It is apparent that certain mechanical

properties and combinations of properties may be obtained in several different classes of alloy steels. In order to indicate more definitely which grades are capable of performing similar service, the following tabulations and comments are given. The automobile is especially suited to this purpose because of the diversity of requirements to be met and the availability of information on quite recent practice. Obviously the list does not include all the important parts of an automobile or all of the steels which have been or might be applied for each part; however, Table XL, taken mainly from references [39] to [43], is a summary of many of the compositions used.

TABLE XL. TYPICAL S.A.E. STEELS USED FOR
AUTOMOBILE PARTS

Part	S.A.E. or Special * Number
1. Balls, rollers, races	4620 4340 6195 52100 SP52100
2. Bumpers	1085 1090 1095
3. Camshafts	*1020* SP1040 (Cast iron)
4. Coil springs	1055 X1055 1060 1065 X1065 SP1065 1090 1095 6150
5. Connecting rods	*1040 1045 X1335 T1335* 3135 SP5128 6135
6. Connecting rod bolts	2330 2350 3130 3145 4130 6135
7. Crankshafts	SP1038 1040 *1045* 3140 3240 4340 (Cast iron)
8. Engine bolts and studs	1035 1040 2330 3130 3140 SP5132 6135
9. Frames	1025 3130
10. Front axles	1040 4130 SP5128
11. King pins	*X1015 X1314 C–Mo* 2315 2520 *3115 4620* 4815 SP5120 6120
12. Leaf springs	1085 1090 1095 SP5148 SP5150 6150 9255 9260
13. Piston pins	*X1020* 1020 *1115 X1314* 2315 2512 3115 *4615* 6115
14. Propeller shafts	SP1030 3130 3140 4140 6135

* SP — Not standard S.A.E. steels — see Table XLI.

TABLE XL. *(Continued)*

Part	S.A.E. or Special Number
15. Rear axle drive gear and pinion	*C–Mo* 2315 2512 3115 *4615* 4820 *6115* SP6122
16. Rear axle shafts	*1040* 1045 *C–Mo* *T1330* 2340 3140 3240 3250 *4140* 4340 SP5145 6140
17. Ring gears	1045 SP1065 *C–Mo* 3115 3120 *5130*
18. Shift and brake levers	1030 1040
19. Steering knuckles	*T1330* *T1340* *C–Mo* 2335 3130 3135 *3140* *4130* *5130*
20. Transmission gears	*C–Mo* 2320 2350 2512 3120 *3145* 3250 3440 4615 *4620* *4640* 4815 *5135* SP5137 SP5145 6115 *6150*
21. Transmission spline shaft	*1060* *C–Mo* 2320 2350 *3115* 3140 3250 4615 *4620* *5135* SP5150

TABLE XLI. STEELS USED BY SIX LEADING AUTOMOBILE MANUFACTURERS

Part	S.A.E. No.	Treatment	Hardness
Camshafts	1020	carburized	75–80 Scl.
	cast alloy	as cast	75 Scl.
Connecting rods	1040	normalized	196 Br.
	1040	quenched and drawn	302 Br.
	1045	normalized	196–217 Br.
	X1335	"	207 Br.
	T1335	"	248 Br.
	SP5128	quenched and drawn	300 Br.
Crankshafts	1045	quenched and drawn	241–248 Br.
	cast alloy	heat treated	269 Br.
King pins	X1015	carburized	60 R " C "
	X1314	"	60 R " C "
	C–Mo	"	File hard
	3115	"	60 R " C "
	4620	"	70 Scl.
	SP5120	"	File hard
	6120	"	File hard
Piston pins	X1020	carburized	60 R " C "
	1115	"	80 Scl.
	X1314	"	60 R " C "
	4615	"	80 Scl.

TABLE XLI. (*Continued*)

Part	S.A.E. No.	Treatment	Hardness
Rear axle drive pinions	C–Mo	carburized	File hard
	4615 or 4620	”	60 R " C "
	6115	”	File hard
	SP6122	”	60 R " C "
Rear axle shaft	1040	quench hardened	302 Br.
	C–Mo	” ”	400–418 Br.
	T1330	” ”	340–363 Br.
	4140	” ”	400–418 Br.
	SP5145	” ”	340–387 Br.
Steering knuckles	T1330–T1340	quenched and drawn	241–302 Br.
	C–Mo	before machining	241–302 Br.
	3140	”	241–302 Br.
	4130	”	241–302 Br.
	5130	”	241–302 Br.
Transmission drive shaft	1060	quenched and drawn	50 R " C "
	C–Mo	carburized	70 Scl.
	3115	”	70 Scl.
	3140	quenched and drawn	50 R " C "
	5135	” ” ”	52 R " C "
	SP5150	” ” ”	52 R " C "
Transmission gears	C–Mo	quenched and drawn	52 R " C "
	3145	” ” ”	52 R " C "
	4620	carburized	58 R " C "
	4640	quenched and drawn	55 R " C "
	SP5137	” ” ”	52 R " C "
	SP5145	” ” ”	52 R " C "
	6150	” ” ”	52 R " C "

COMPOSITIONS OF " SP " OR SPECIAL STEELS SHOWING TYPICAL DEVIATIONS FROM STANDARD S.A.E. ANALYSES [42]

	C	Mn	Si	Cr	V
SP1030	0.27–0.35	0.70–0.90	0.07–0.15	—	—
SP1040	0.38–0.42	0.70–0.90	0.07–0.15	—	—
SP1065	0.60–0.70	0.70–0.85	0.15–0.20	—	—
SP5120	0.18–0.22	0.65–0.75	0.10–0.20	0.80–0.95	—
SP5128	0.26–0.30	0.65–0.80	0.10–0.20	0.80–1.00	—
SP5137	0.35–0.38	0.65–0.80	0.10–0.20	0.90–1.10	—
SP5145	0.42–0.47	0.70–0.90	0.10–0.20	0.85–1.10	—
SP5150	0.48–0.52	0.70–0.90	0.10–0.20	0.85–1.10	—
SP6122	0.20–0.24	0.60–0.75	0.10–0.15	0.65–0.80	0.12–0.15

Courtesy of W. H. Graves and J. L. McCloud, " Metal Progress." [40, 42]

The steels which are in bold face type in Table XL were discussed by Graves.[40]   They represent the 1936 practice of six leading automobile manufacturers.   Further information on the type of heat treatment used and hardness requirements for these steels is included in Table XLI.

The steels marked " SP " in Tables XL and XLI were used by a large manufacturer in 1931, and although many important changes in practice have been made in recent years, they are included primarily to show typical deviations from standard S.A.E. analyses, including narrower carbon ranges as well as changes in alloy requirements.   The special or " SP " designations were added to the nominal S.A.E. numbers by the author for convenience in comparing the modified composition limits given in Table XLI with the basic S.A.E. compositions.   It is common practice for the users of S.A.E. steels to modify the composition ranges slightly in their specifications.

The " C–Mo " designations refer to the carbon-molybdenum steels used extensively by one manufacturer ("Amola" steels).   The composition ranges of these steels are: 0.20–0.70% C, 0.20–0.30% Si, 0.70–0.90% Mn, and 0.15–0.25% Mo.   The carbon contents are similar to those of other steels used for the same parts.

It will be noted that the hardness requirements for a given part are in fair agreement except for those parts where there is a choice between carburizing and direct hardening practice.   In such cases the surface of the carburized part will be harder and its interior softer than a non-carburized part for the same application.   When high surface hardness is required on a non-carburizing steel, such as

TABLE XLII.   TYPICAL FERROUS METALS USED IN AIRPLANE
ENGINE CONSTRUCTION

Application	S.A.E. No.	Tensile Strength, lb./sq. in.	Yield Strength, lb./sq. in.	% El. in 2 in.	% Red. in Area
Ferrules, clips, lock wire	1010	38 000	20 000	35	—
Camshaft, washers, ball ends	1015	45 000	25 000	22	—
Nuts, screws, counter-weights, flanges	1020	55 000	30 000	22	—
Shafts, sleeves, nuts, rivets	1035	80 000	50 000	20	—

TABLE XLII. (*Continued*)

Application	S.A.E. No.	Tensile Strength, lb./sq. in.	Yield Strength, lb./sq. in.	% El. in 2 in.	% Red. in Area
Cylinder barrels, keys ..	1050	100 000	75 000	16	—
Springs	1090	225 000 to 350 000	—	—	—
Shims, wearing parts of valve mechanism	1095	—	—	—	—
Screws, nuts, dowels for minor attachments ..	1120	55 000	—	—	—
Bolts, studs, nuts, shafts	2330	125 000	100 000	15	50
Connecting rods, gears .	2340	135 000	110 000	15	50
Gears, piston pins	3115	130 000	90 000	16	40
Bolts, studs, shafts	X3140	130 000	100 000	17	50
Crankshaft, drive shafts	3240	135 000	110 000	15	50
Gears, pins	3250	225 000	200 000	10	40
Gears, cams, crankshaft	2515	170 000	145 000	14	45
Gears, drive shafts, cams	3312	160 000	135 000	15	50
Washers, shims, spacers, tubes	X4130	95 000	75 000	12	—
Crankshaft, connecting rods	4340	160 000	140 000	15	50
Balls, bearings, knuckle pins	52100	—	—	—	—
Gears, shafts, propeller hubs	6135	150 000	120 000	15	50
Piston pins, gears, drive shafts, tappets, springs	6150	220 000	200 000	10	—
Ball ends, dowels, tappets, bolts, studs	6190	—	—	—	—
Gears, cylinders	Nitralloy [a]	135 000	100 000	18	55
Piston pins, shafts, pump liners, bushings	" [b]	120 000	80 000	15	45
Exhaust manifolds, supercharger casing	30905 [c]	100 000	35 000	40	—
Valve, inlet	71360	55 000 [d]	—	—	—
Valve, inlet tips	71665	60 000 [d]	—	—	—

[a] Composition of Nitralloy: 0.43% C, 0.55% Mn, 1.6% Cr, 0.4% Mo, 1.2% Al.
[b] Composition of Nitralloy: 0.35% C, 0.45% Mn, 1.2% Cr, 0.2% Mo, 1.2% Al.
[c] Composition of 30905: 0.07% C, 0.45% Mn, 17% Cr, 7% Ni, 0.2% Ti.
[d] Tensile strength at 1200° F. — All other tensile properties are typical specification values.

Courtesy of J. B. Johnson, condensed from Trans. Society of Automotive Engineers.[44]

TABLE XLIII.   PRINCIPAL AIRPLANE ENGINE PARTS AND REPRESENTATIVE SPECIFICATIONS

Name of Part	Spec. No.	Hardness	Spec. No.	Hardness	Spec. No.	Hardness	Spec. No.	Hardness
Cylinder barrels	1050	225	4140	300	Nitralloy	900	—	—
Valve, spring	1095	C-44	6150	C-44	—	—	—	—
Valve, spring washer	3135	C-35	6150	C-40	—	—	—	—
Valve, spring retainer	3250	C-50	6150	C-40	—	—	—	—
Rocker arm	3140	C-30	6150	C-30	2330	C-30	—	—
Rocker arm hub bolt	3140	C-32	6150	C-25	3312	C-60	—	—
Rocker arm cup	10115	C-62	3250	C-42	52100	C-60	6180	C-55
Push rod, ball end	1015	C-62	3250	C-46	1095	C-50	—	—
Push rod, roller	3115	C-60	3215	C-60	—	—	—	—
Push rod, roller pin	3140	C-40	3250	C-50	6150	C-52	—	—
Cam	3250	C-55	2515	C-60	—	—	—	—
Camshaft	1015	C-60	—	—	—	—	—	—
Cam drive shaft and gear	3312	C-60	3140	C-38	2515	C-55	—	—
Impeller shaft	2515	C-60	3312	C-60	Nitralloy	900	—	—
Crankshaft	X3140	260	3240	280	4340	320	2515	C-60
Crankshaft extension	3140	C-40	3250	C-45	2515	C-55	—	—
Propeller hub nut	6135	250	3312	C-60	2330	C-30	—	—
Rod, connecting	3140	270	4340	350	2340	340	—	—
Pin, piston	6150	C-50	3312	C-60	3250	C-47	Nitralloy	900
Pin, knuckle	3312	C-60	3120	C-60	52100	C-60	—	—
Bolts, conn. rod, crankcase	3250	C-35	6150	C-35	3140	C-30	—	—
Studs, cylinder	6150	C-26	3140	C-30	—	—	—	—
Nuts	6150	B-95	3140	C-25	2330	C-20	—	—
Gears, reduction; camshaft drive; accessory drive	3250	C-45	3312	C-60	2515	C-60	Nitralloy	900

Hardness values are taken from service parts and are Brinell (10 mm. ball-3000 kg. load) unless preceded by C or B for Rockwell-Cone and 1/16 ball, respectively. Hardness of nitrided Nitralloy parts determined by Vickers-Brinell.
Courtesy of J. B. Johnson, condensed from Trans. Society of Automotive Engineers.[44]

S.A.E. 6150 for transmission gears, it may be quenched from a cyanide bath which supplies 0.001 inch to 0.003 inch of very hard skin and eliminates the possibility of a soft decarburized surface. (See Chapter X.)

Steels Used for Airplane Engines. — Typical practice for airplane engine materials has been summarized by J. B. Johnson.[44] Although the aircraft industry has followed automotive practice to a considerable extent in selecting steels, great emphasis has naturally been placed on safe materials very carefully processed, finished, and tested, with the idea of attaining a high ratio of strength to weight always in mind.

According to Johnson the material for any particular engine part is selected on the basis of the following factors which are listed in order of importance.

(1) Mechanical properties at the operating temperature.
(2) Corrosion resistance for parts in contact with fuel or the products of combustion.
(3) Uniformity, freedom from soft spots, heat-treatment cracks and magnaflux indications.[45]
(4) Suitability for fabrication by the methods and equipment available to the manufacturer.
(5) Machinability.
(6) Cost.

Tables XLII and XLIII give representative data on the type compositions used and some of the property specifications. A large number of nonferrous and high alloy parts have been omitted in this condensed form but may be found in the original article.

REFERENCES

1. Hildorf, W. G., and McCollam, C. H., "Classifying Steels by Sparking," Metal Progress, **23**, 23 (Feb. 1933)
2. Bain, E. C., "The Alloying Elements in Steel," Metals Handbook, 519, Am. Soc. for Metals, Cleveland (1939)
3. Zimmerman, J. G., Aborn, R. H., and Bain, E. C., "Some Effects of Small Additions of Vanadium to Eutectoid Steel," Trans. Am. Soc. for Metals, **25**, 755 (1937)
4. Priestley, W. J., "Metals and Ferroalloys Used in the Manufacture of Steel," Metals Handbook, 803, Am. Soc. for Metals, Cleveland (1939)
5. Dawe, C. N., "Steel Physical Properties Atlas," Am. Soc. for Metals, Cleveland (1936)
6. Janitzky, E. J., and Baeyertz, M., "Marked Similarity in Tensile Properties of Sev-

eral Heat Treated S.A.E. Steels," Metals Handbook, 515, Am. Soc. for Metals, Cleveland (1939)

7. Boegehold, A. L., "Plain Carbon Steel, Its Utility for Automobiles," Metal Progress, **31,** 147 (1937)

8. Stoughton, B., and Harvey, W. E., "The Relative Merits of Some Different Alloy Steels With Respect to Certain Mechanical Properties," Proc. Am. Soc. for Testing Mats., **30,** Part II, 241 (1930)

9. Boegehold, A. L., "Appraising a Steel for a Given Duty," Metal Progress, **31,** 403 (1937)

10. Knowlton, H. B., "Physical Properties of Axle Shafts," Trans. Am. Soc. for Metals, **25,** 260 (1937)

11. Scott, H., and Hoop, J. G., "The Hardening Transformation in Manganese Steels," Trans. Am. Soc. for Steel Treating, **21,** 233 (1933)

12. Krivobok, V. N., and Wells, C., "Constitution of Iron-Manganese Alloys," Metals Handbook, 380, Am. Soc. for Metals, Cleveland (1939)

13. Thum, E. E., "The New Manganese Alloy Steels," Proc. Am. Soc. for Testing Mats., **30,** Part II, 215 (1930)

14. Fulton, A. O., "Manganese As An Alloying Element in Pearlitic Steels," Metals Handbook, 561, Am. Soc. for Metals, Cleveland (1939)

15. Austin, C. R., "Observations on the Effect of Normalizing Medium Manganese Steels on the Microstructure and Physical Properties," Trans. Am. Soc. for Steel Treating, **21,** 435 (1933)

16. Ruder, W. E., "Silicon As An Alloying Element in Steel," Metals Handbook, 586, Am. Soc. for Metals, Cleveland (1939)

17. Greiner, E. S., Marsh, J. S., and Stoughton, B., "Alloys of Iron and Silicon," The Engineering Foundation, McGraw-Hill Book Company, New York (1933)

18. Herzig, A. J., and Parke, R. M., "Low Temperature Impact Properties of Some S.A.E. Steels," Metals and Alloys, **9,** 90 (1938)

19. Wickenden, T. H., "Nickel As An Alloying Element in Steel," Metals Handbook, 577, Am. Soc. for Metals, Cleveland (1939)

20. Waterhouse, G. B., "Chromium Rail Steels," Metal Progress, **25,** 32 (April 1934)

21. Kinzel, A. B., and Crafts, W., "The Alloys of Iron and Chromium, Vol. I — Low-Chromium Alloys," The Engineering Foundation, McGraw-Hill Book Company, New York (1937)

22. Miller, W. B., "Properties of Cromansil Steels for Structural Purposes," Steel, **94,** 32 (May 21, 1934)

23. Kinzel, A. B., "Chromium and Manganese in Structural and Stainless," Metal Progress, **29,** 46 (March 1936)

24. Roberts, G., "A New High Tensile Steel for Structural Work," Structural Engineer, **12,** 314 (1934)

25. Chandler, H. T., and Dawe, C. N., "Develops Manganese-Chrome-Vanadium Steel for Automotive Forgings," Steel, **93,** 23 (Sept. 25, 1933)

26. "Advancement in Low and Medium Alloy Steels," Metal Progress, **26,** 5 (Oct. 1934)

27. "Low Alloy, High Yield Strength Structural Steels — An Extended Abstract," Metals and Alloys, **7,** 77 (1936)

28. Epstein, S., Nead, J. H., and Halley, J. W., "Choosing a Composition for Low-

alloy High-strength Steel," Trans. Am. Inst. Mining and Met. Eng., **120,** 309 (1936)

29. Cone, E. F., " The Present Status of the Low Alloy High Strength Steels — A Survey," Metals and Alloys, **9,** 243 (1938)

30. Stuebing, A. F., " High Tensile Steels for Light Weight Construction," Metal Progress, **34,** 37 (1938)

31. Bullens, D. K., and Metallurgical Staff of Battelle Memorial Institute, " Steel and Its Heat Treatment," Vol. II, John Wiley and Sons, New York (1939)

32. Gregg, J. L., and Daniloff, B. N., " The Alloys of Iron and Copper," The Engineering Foundation, McGraw-Hill Book Company, New York (1934)

33. Thum, E. E., " Izod Toughness vs. Tensile Strength of Heat Treated S.A.E. Steels," Metal Progress, **29,** 57 (May 1936)

34. Goldstein, J. R., " Metals Used in the Aircraft Industry," Metal Progress, **33,** 357 (1938)

35. Gregg, J. L., " The Alloys of Iron and Molybdenum," The Engineering Foundation, McGraw-Hill Book Company, New York (1932)

36. Clark, C. L., and White, A. E., " Creep Characteristics of Metals," Trans. Am. Soc. for Metals, **24,** 831 (1936)

37. Kanter, J. J., " Interpretation and Use of Creep Results," Trans. Am. Soc. for Metals, **24,** 870 (1936)

38. Priestley, W. J., " Vanadium in Steel and Iron," Metals and Alloys, **9,** 193 and 237 (1938)

39. Boegehold, A. L., " Advantages of Alloys Steels," Metal Progress, **31,** 265 (1937)

40. Graves, W. H., " What Steel is Best for This or That Part?", Metal Progress, **29,** 36 (April 1936)

41. " S.A.E. Handbook," Soc. of Automotive Engineers, New York (1938)

42. McCloud, J. L., " Steels Used in Ford Industries," Metal Progress, **19,** 32 (March 1931)

43. " Nickel Alloy Steel, Bulletin U–1," The International Nickel Company, Inc. (1936)

44. Johnson, J. B., " Aircraft Engine Materials," Trans. Soc. of Automotive Engineers, **32,** 153 (1937)

45. Noble, H. J., " Magnaflux Inspection Methods in Airplane Engines," Metals and Alloys, **7,** 167 (1936)

SURFACE TREATMENT OF STEEL

The surface of steel may be modified to improve its appearance, to increase its hardness and wear resistance, to improve corrosion or heat resistance, or to prepare the metal for painting. A wide variety of commercial processes are in use ranging from galvanizing to nitriding. It will be possible to treat only the most important processes here, with special emphasis on carburizing and other methods of increasing wear resistance. In addition to the specific processes outlined in Table XLIV, many other protective and ornamental coatings are applied by electroplating and by the molten metal spray gun.[1]

TABLE XLIV. COMMERCIAL PROCESSES FOR TREATING THE SURFACE OF STEEL

Process	Nature of Coating and Method of Application
Galvanizing	Applied by dipping wire, sheet, or finished products into molten zinc.[1,2] Wire is now zinc coated by electrodeposition as well as by hot dipping.[3]
Sherardizing	Articles are treated with zinc dust in a sealed rotating drum at a temperature of 700° F.[4,5]
Terne coating	Sheets are coated by immersion in a molten bath usually containing from 12 to 40% tin, balance lead.[6]
Tin coating	Sheets or formed articles are coated in a bath of pure molten tin.[6] Sheets are now being tinned by electrodeposition as well as by hot dipping.
Cadmium plating	Applied by electrodeposition.[1,4,7]
Calorizing	A coating of high aluminum content is produced by heating steel or iron objects in a sealed rotating drum at a temperature of about 1700° F. using a mixture of powdered aluminum, aluminum oxide, and ammonium chloride, and using an inert or reducing atmosphere. After removal from the drum a high temperature diffusion heat treatment may be

TABLE XLIV. (*Continued*)

Process	Nature of Coating and Method of Application
	applied. The coating consists mainly of an aluminum-iron intermetallic compound. There are several hot dip processes for applying aluminum coatings and the metal spray gun is also used.[1, 4, 8, 9]
Parkerizing	An iron phosphate coating is applied by immersing iron or steel in a hot solution of manganese dihydrogen phosphate for 30 to 60 minutes.[1, 4] The coated articles are gray in color but become black after oiling or waxing.
Bonderizing	A series of modifications of the Parkerizing process in which thinner phosphate coatings are applied in from thirty seconds to ten minutes by spraying or dipping in hot phosphate solutions containing catalysers. Several similar processes have been developed using zinc-phosphate, zinc-nitrate, oxalic acid, chromic acid and sodium chloride as the reactive agents and, in some cases, applying the coating by electrolysis.[1, 4] These treatments are intended as a base for painting and give only temporary protection unless covered by a paint film.
Steam bluing	A protective coating of Fe_3O_4 is formed by applying steam while the steel is at a dull red heat. The original blue color is darkened if oil or wax is applied for protection against moisture. (Tieman[9] and Young[10] describe gun metal finish, browning, and other coloring processes.)
Porcelain enameling	A fine powdered frit of borosilicate glass is applied to iron or steel objects in very thin coatings by either spraying or dipping. The glass is then fired at temperatures near 1500° to 1600° F. to form an adherent coating. When applied to cast irons the coatings are much thicker than on wrought sheet metals and the firing temperatures are lower.[1, 11]
Siliconizing	Silicon is introduced into the surface of iron or steel by treatment with silicon carbide and chlorine at temperatures between 1700° and 1850° F.[12]
Chromizing	Chromium is diffused into the surface of steel by heating to high temperatures in contact with powdered alumina and chromium in a protective atmosphere such as hydrogen.[13, 14] Other methods include diffusion at high temperatures of electrodeposited chromium, and cementation in a mixture of chromium oxide, chromium chloride, and powdered aluminum.[14]

TABLE XLIV. *(Continued)*

Process	Nature of Coating and Method of Application
Carburizing	Carbon is introduced by means of carbon monoxide or hydrocarbon gases at temperatures between 1550° and 1750° F. Molten salt baths are also used as carburizing media.
Cyaniding	Carbon and nitrogen are introduced by heating in molten cyanides, usually at temperatures between 1500° and 1600° F.
Nitriding	Nitrogen is introduced by heating in partially dissociated ammonia gas at 900° to 1100° F.

Zinc Coatings. — Zinc coatings were among the first used for the protection of steel from atmospheric corrosion, and galvanizing will probably remain one of the principal methods for prolonging the life of iron and steel products. A protective coating of corrosion products forms on the zinc under normal atmospheric conditions. One of the most desirable features of galvanized coatings is the comparative freedom from local attack at seams, edges, and places where the coating is broken due to abrasion, peeling, pitting, or coating defects. Under these circumstances the anodic or electronegative condition of zinc with respect to iron promotes general solution of the coating rather than the steel.

Irregular shaped objects, structural members, and formed products such as pails and tubs are usually hand dipped. Bolts, nuts, and other small steel parts may be given a uniform zinc coating by sherardizing.[5] Sheets and wire are usually machine dipped. Machine dipping ordinarily involves proper preparation of the steel by pickling and washing, passage through a flux of ammonium chloride and zinc chloride into a bath of molten zinc held at about 840° F., and removal from the bath through rolls or wiping devices which aid in controlling the thickness of the coating.[1, 2, 15] The speed at which the metal passes through the bath and the temperature of the bath are also important in controlling the thickness of the coating. Zinc coatings are made of various weights, usually from 0.4 to 2.5 ounces per square foot of surface. The resistance to corrosion increases approximately as the weight of the coating increases.

Tight hot dipped coatings which will stand relatively severe form-

ing operations without cracking or peeling are necessarily thin. In sheet galvanizing practice, tight coatings ordinarily carry less than 1 ounce of zinc per square foot of sheet. (Specifications for galvanized sheets are based on sheet area rather than surface area.) Moderately heavy coatings, $1\frac{1}{4}$ to 2 ounces, are applied to sheets 30 gage or heavier and will stand light draws and bends. Heavy coatings, 2 to $2\frac{1}{2}$ ounces, are applied to 24 gage and heavier and will stand corrugation and large radius bending as in forming culvert pipes. Extra heavy coatings, $2\frac{3}{4}$ ounces, are applied to 18 gage and heavier and will stand corrugating only. (See A.S.T.M. Specification A93–27 for further information, including tests for weight of coating.)

The size, shape, and brightness of the zinc spangle can be controlled in processing. Dull finishes, for example, are produced by high bath temperatures and slow cooling after coating.

One of the difficulties in the handling of galvanized sheets is the production of a white stain consisting of zinc oxide and zinc hydroxide. Upon normal exposure to the atmosphere these corrosion products form an invisible protective film which retards further action. When the supply of air is restricted, as in piles of stored sheets, and when moisture is present, the formation of corrosion products sometimes proceeds very rapidly. Heated warehouses are necessary for safe storage in moist climates.

The microstructure of a hot dipped coating generally consists of three constituents arranged in layers as in Fig. 153. In the hot dip process the time for alloying of the zinc and iron is short; nevertheless, a considerable amount of a zinc-rich compound or constituent and a smaller amount of a compound * higher in iron are usually formed. The latter compound, often referred to as the primary alloy, is shown as a thin irregular layer next to the steel base in Fig. 153. The exact compositions of these compounds cannot be determined with certainty by examination of the zinc-iron equilibrium diagram because they do not attain equilibrium during their short period of formation. Furthermore, the relationships of the binary

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* Although these microconstituents are essentially intermetallic compounds, their compositions and formulas are somewhat uncertain and probably complicated by the presence of zinc in solid solution.

FIG. 153. — HOT DIPPED GALVANIZED COATING ON STEEL.   (Etching reagent: one drop of a stock solution of 96 ml. HNO_3 + 4 g. CrO_3 to 20 ml. of alcohol.) X1000.

FIG. 154. — GALVANNEAL COATING ON STEEL.   (Etched the same as Fig. 153.) X1000.
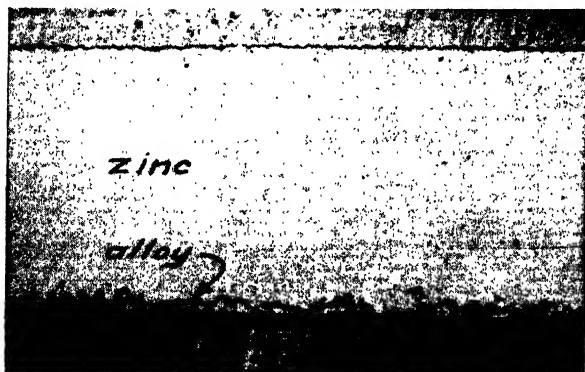
FIG. 155. — ZINCGRIP COATING ON STEEL.   (Etched the same as Fig. 153.)   X1000.

zinc-iron system are modified by the presence, in certain types of coatings, of up to about 1.5% tin and 1% lead as well as other elements which may be added or present as impurities.

Not only the chemical compositions but the mechanical properties of the alloy layers may be varied in different types of coatings. The outstanding characteristics of the intermetallic compounds are their inherent brittleness and their capacity for forming a tight bond with the base metal.

Special Galvanizing Processes. — One of the major problems in galvanizing is the production of a coating which will withstand forming operations without cracking or peeling. The bond between the coating and the steel may be improved by several methods including heating of the coated metal, either before or after it has cooled from the galvanizing treatment, thereby promoting the formation of intermetallic compounds of iron and zinc at the interface. In the " galvanneal " process the coated product is maintained at a temperature near or above the melting point of the zinc until the entire coating is thus converted.[15] The structure of such a coating having two compound layers is shown in Fig. 154.

It has been shown that although coatings consisting of iron-zinc compounds become severely cracked in forming, they do not peel off in large flakes but crumble into minute fragments which tend to adhere to the base metal. While this condition cannot be said to improve corrosion resistance, the influence of microscopic cracks in zinc coatings is not nearly as great as in nickel or chromium electroplated coatings in which rapid corrosion of the steel may occur by galvanic action.

The " galvanneal " product is not spangled and the formation of intermetallic compounds gives a mottled surface which is much superior to a smooth zinc surface in the initial adherence of paints. Exposure of hot dipped coatings to the atmosphere, chemical etching, or mechanical roughening will also improve the initial adherence of paint; however, subsequent reaction between the coating and the paint tends to produce failure unless a nonreactive surface is provided before painting. In addition to the use of special paint primers, ordinary hot dipped coatings are now treated chemically by a modified Bonderizing process to form a stable surface which is

very suitable for painting, giving good initial adherence and long life.

Sheets and wire for forming operations can now be coated by the " zincgrip " process, in which it is possible to apply a heavy coating of relatively pure and ductile zinc firmly bonded to the steel by a thin alloy layer, as illustrated in Fig. 155. So-called " tight " coatings made by the hot dip process are restricted to relatively thin layers having limited corrosion life. Recent developments such as " zinc-grip " and electrogalvanized sheets are designed to meet the requirements of a heavily coated product capable of withstanding difficult fabricating operations without peeling or powdering.

Terne Plates and Long Ternes. — Terne plates, or short ternes, are light gage alloy coated sheets used extensively for roofing. They are usually sold in the standard size 20″ x 28″, and in units known as the " base box " containing a fixed area of plate (31,360 sq. in., the equivalent of 56 standard sheets). The lead-tin alloy coatings used contain about 25 % tin and range from 8 to 40 pounds of alloy per double base box.

Long ternes are made in any ordinary size sheet or strip. The steels coated are likewise more variable than for short ternes and include drawing and deep drawing grades for many special applications. Long ternes are also used for building purposes including roofing, eaves troughs, conductor pipe, window frames, and doors. Certain grades of long ternes are made with very smooth finishes suitable for painting. Terne coat retains paint very satisfactorily and prolongs the life of the paint coating. Still another grade is used for containers for lubricating oil and other products exclusive of foods. Very thin coatings, comparable to tin plate coatings, may be used for such purposes.

The lubricating effect of the high lead alloy coating makes terne sheets particularly desirable for die forming operations. The formed products are very easy to solder, which is one of the important advantages of lead alloy coatings. Typical formed products are caskets, automobile gasoline tanks, oil filters, and radiator core parts.

The drawing grades of terne coated sheets are often called alloy coated sheets. They have a light weight coating, about 8 pounds

per double base box, and the coating is intended to give protection against rusting only during the period of storage and processing. On the other hand, roofing ternes with 40 pounds of alloy per double base box depend upon the thick coating to protect the steel surface in service.

Tin Plate. — Tin plate [6] for cans and other food containers is an important item in steel production. No cheaper substitute for tin has proved entirely satisfactory as a protective coating for resisting food acids.

According to McKay and Worthington [7]: ". . . the problem of tin can corrosion centers around the facts that tin is not corroded by foodstuffs in the absence of oxidizers but is corroded more or less in their presence, that iron is corroded in either the absence or the presence of oxidizers, and that iron is cathodic to tin under the conditions of the tin can. . . .

" The tin can is successful where the amount or availability of the oxidizer that may be in the foodstuff is insufficient to strip the thin layer of tin from the iron. The iron exposed at imperfections in the tin coating is of such minute area that, cathodic as it is to the tin, it remains unattacked. Furthermore, the sacrificial corrosion of tin in consequence of its couple with a minute area of iron is immaterially small."

The polarity of the iron-tin couple is generally reversed when the metal surface is aerated, iron being attacked preferentially; furthermore, tin itself is corroded more rapidly under such conditions, therefore heavy tin coatings are used for milk containers, tin coated cooking utensils, and similar products.

Although tin coatings are very ductile, adhere well to steel, may be readily soldered, and are pleasing in appearance, their relatively high cost prevents their use in competition with galvanizing and terne coat for general application outside the field of food containers.

Information on the manufacture of tin plate and other coated sheets will be found in " The Making, Shaping, and Treating of Steel." [10] The coating processes and the characteristics of coated metals are described in " Protective Coatings for Metals." [1] Chapters VII, IX, and X of " Corrosion Resistance of Metals and Al-

loys " [7] will also be of interest for a further study of the corrosion resistance of tin, zinc, and lead coatings.

Cadmium Plating. — Cadmium is similar to zinc in its capacity to protect iron and steel from corrosion; however, with the possible exception of exposure to marine atmospheres, corrosion rates for cadmium are higher than those for zinc under similar conditions. Although cadmium is applied by electroplating, cadmium coatings are included here because they are more closely related to galvanized and other soft metal coatings than to electrodeposits such as nickel or chromium. The principal applications of cadmium plating are screws, bolts, and other small hardware articles for both indoor and outdoor use, and for aircraft and marine fittings.

Aluminum Coatings. — Aluminum coatings applied by calorizing and similar processes resist scaling at elevated temperatures up to about 1800° F. and are used for many heat resisting applications at lower temperatures. Aluminum coatings for both heat and corrosion resistance contain less of the brittle iron-aluminum intermetallic compound when applied by hot dip processes. The spray gun method of application is used principally for assembled equipment or structures. As in other applications of aluminum, the superficial oxide film which forms at the surface is the basis of its protective action.

Porcelain Enamel. — Porcelain or vitreous enamel is assuming increasing importance as a coating for steels used in household fixtures and appliances, processing equipment, and steel buildings. It has the characteristics of a glass rather than those of a paint or low temperature baked lacquer, therefore its limitations are cracking and checking due to thermal and mechanical shock, modified, however, by the supporting action of the metal.

While vitreous enameled cast iron has been in use for many years for bathtubs and other plumbing fixtures, enameling sheets of very low carbon steels and ingot iron have extended this field to formed sheet metal products such as washing machine tubs, refrigerator and stove cabinets, building and sign panels. Cooking utensils have been made from enameled steel for many years. More recently domestic water heaters, large vats, pipe lines, and processing equipment for the chemical, food, and liquor industries have been made of steel or cast iron with heavy porcelain linings. The attractive and perma-

nent colors which are available make porcelain enamel highly desirable as a finish on products for both household and outdoor use.

The microstructure of a typical porcelain enamel coating is shown in Fig. 156. The dark layer near the surface of the steel is known as the ground coat. During firing it penetrates the surface of the steel, which is always given a rough finish, forming the close bond observed in the photomicrograph. The white or colored cover coats fuse into the ground coat upon subsequent firings. The bubbles
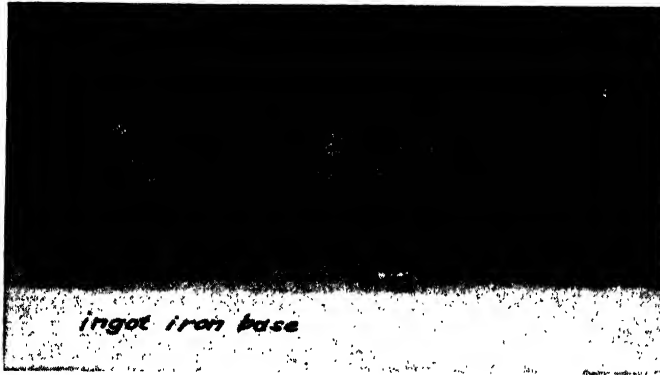
FIG. 156. — PORCELAIN ENAMEL COATING ON INGOT IRON. (Unetched.) X100. (*Courtesy of George E. Koeppel.*)

shown in Fig. 156 are similar to those found in all vitreous enamels. It is considered that they improve the elasticity of the coating.

Phosphate Coatings. — Bolts and screws and many other small parts are Parkerized for protection against corrosion; however, the original Parkerizing process has limited application. The thinner Bonderized coatings and similar treatments are now used extensively as a base for painting, lacquering, and enameling, especially in the automotive industry where fenders, hoods, headlights, wheels, and other sheet steel parts are treated before painting.[4] Refrigerator cabinets, burial vaults, license plates, and typewriter frames are other representative uses. The low temperature baked enamels which are frequently applied on these parts should not be confused with the porcelain enamels which are fired at a red heat.

A modification of the Bonderizing treatment has been developed for use on galvanized sheets as a base for painting. (See page 319.)

The application of a phosphate coating to piston rings, pistons, camshafts, and other machine parts to eliminate " scuffing " during the wearing-in period has recently been reported.[17] The coating is applied by dipping in a hot acid bath. It increases the " oiliness " (or capacity for being wet by the oil) rather than the superficial hardness of the part.

Oxide Coatings. — Steam bluing and its modifications may be applied to steels for the purpose of obtaining color effects and, to a limited extent, protection from corrosion. Oxidation may be effected by heat alone, and many other colored finishes may be produced by special salt baths, oil, chemicals, and electrolytic treatments.

Siliconizing. — Silicon impregnation by a new patented process known as " Ihrigizing " has recently been commercialized.[12] A case depth of 0.025–0.030 inch is produced on low carbon steels in two hours. Although the case is not exceptionally hard (Rockwell " B " 80 to 85), it is not machinable and working surfaces are finished by grinding.

The silicon content of the case is quite uniform, varying from about 14.0 to 13.5%, which is approximately the same as the silicon content of certain alloy cast irons which are well known for their excellent resistance to many strong acids. (See page 468.) As applied to low carbon steels, the Ihrigizing process offers the advantage of a tough core and a fairly ductile and very adherent case, whereas the high silicon castings are brittle. Malleable iron fittings and other iron and steel products have been treated successfully by this process, but low carbon low sulphur steel is recommended for best results.

In addition to good corrosion resistance in hot nitric, sulphuric, hydrochloric, and other acids, silicon impregnated steels are resistant to scaling at high temperatures, and have good wear resistance for applications such as automotive water pump shafts.

Chromizing. — Chromizing is a relatively expensive process of limited application. Carbon steels with under 0.20% carbon are the most satisfactory base. A penetration of 0.01 inch may be obtained at 2375° F. in about 4 hours.[14] Lower temperatures give even slower penetration. According to Kelley [14] the coating con-

tains 10 to 20% chromium, thus a chromized surface would be expected to have resistance to oxidation and corrosion comparable to that of certain grades of stainless steels. (See Chapter XIV.)

Metal Spray Process. — The molten metal spray gun, mentioned earlier in this chapter, is undoubtedly a most versatile medium for applying metal coatings. The most widely used instrument for metal spraying is essentially a gas fired pistol in which wire of the composition to be deposited is fed continuously into an oxyacetylene flame in which it is melted and then sprayed from the gun in a finely divided form by means of an air blast. Powdered metal is used, instead of wire, in another process for metal spraying. The impact of the metal particles is usually sufficient to give good mechanical adherence to a surface properly prepared by sandblasting or other means; however, the bonding effect of solid solution or compound formation with the base metal is lacking unless the object is subjected to a diffusion heat treatment after coating. The nature of the coating itself is described by R. M. Burns [18]: " The minute semimolten particles which comprise the spray acquire a superficial oxide coating which becomes incorporated in the coating. The particles of metal upon impinging upon the base metal are flattened into relatively thin scales which are rapidly chilled by the passing air blast. As the coating is built up these scales become interlocked with others preceding and following them, giving rise to a structure of rather high intrinsic tenacity. There are, however, cavities of microscopic size between the scales at numerous points with the result that the coating is somewhat porous in nature. These cavities, together with small amounts of oxide, have the effect of reducing the density of sprayed metals." (Data given by Burns for sprayed coatings of aluminum, zinc, tin, copper, lead, brass, and bronze indicate densities of 85 to 95% of corresponding cast metals.) Despite their reduced densities, the scratch hardness values for the sprayed coatings are generally higher than for castings.

Compared with hot dipped or electrodeposited coatings, somewhat thicker deposits of sprayed coatings are required for comparable protection of iron or steel from corrosion. This is especially true in the case of tin, nickel, and other metals which are electroposi-

tive (cathodic) to iron and depend, therefore, on complete coverage for protection against corrosion.

Of the sprayed metal coatings used to protect steel structures from corrosion, zinc is the most important. Sprayed aluminum coatings are used for heat resisting applications such as combustion chambers of engines, exhaust pipes, and permanent metal molds.

Although the lower melting metals and alloys are most readily handled in a spray gun, carbon steels may also be deposited. One of the most useful fields for both ferrous and nonferrous deposits is building up worn bearing surfaces of shafts, cylinders, valves, rolls, and other machine parts.

Carburizing. — Carburizing is one of the oldest heat treating processes. Some of the early sword blades and tools are believed to have been made by the cementation or carburization of low carbon wrought irons. In modern times the cementation process was the first to produce steel in commercial quantities. Since carburization depends upon the ability of iron to absorb carbon at red heats, steels produced by this method required prolonged heating at high temperatures in the presence of carbonaceous gases. The iron to be carburized was generally packed in charcoal in a closed container. Cemented or " blister " steel was produced in England over a period of about one hundred years from 1750 to 1850.

Present day carburizing practice has surface hardening as its object. Soft, easily forged, and easily machined low carbon steels are given a hard tool steel surface by heating for a few hours in a carburizing atmosphere. The resulting product has the toughness of a low carbon steel combined with the wear resistance of a tool steel.

Practically all phases of carburizing have recently been reviewed in the American Society for Metals " Carburizing Symposium." [19]

Mechanism of Carburization. — Carbon monoxide is believed to be the active carrier of carbon to the surface of the steel in " pack " carburizing.[20, 21, 22] The reactions at the surface may be expressed as:

$$2CO \rightleftharpoons CO_2 + C \qquad \text{and}$$
$$2CO + 3Fe \rightleftharpoons CO_2 + Fe_3C$$
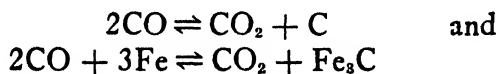
If we accept the hypothesis that carbon is soluble in austenite only as elemental carbon, the first reaction applies. If cementite forms directly according to the second equation, it is immediately soluble in gamma iron (which is stable at the carburizing temperatures); thus it loses its density as a compound until the temperature is lowered.

In " gas " carburizing, using natural or manufactured gas, the hydrocarbons methane and ethane are the principal sources of carbon. The carburizing reactions for methane may be represented by the equations:

$$CH_4 \rightleftharpoons 2H_2 + C \qquad \text{and}$$
$$CH_4 + 3Fe \rightleftharpoons 2H_2 + Fe_3C$$
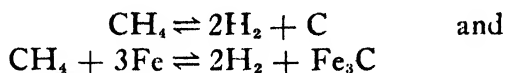
Once the carbon is deposited at the surface as Fe_3C or as elemental carbon in solid solution, it tends to diffuse inward towards zones of lower carbon concentration.[23] Provided a sufficiently high concentration of carbon is maintained at the surface, the depth of case depends on the rate of diffusion of carbon (or Fe_3C) through the iron. This rate increases with temperature. Data reported by Schlumpf[22] for carburization of S.A.E. 3115 in a vertical retort using natural gas include the following case depths obtained in four hours at the indicated temperatures: 1550° F. — 0.024 in.; 1600° F. — 0.029 in.; 1650° F. — 0.036 in.; 1700° F. — 0.047 in.; 1750° F. — 0.063 in. Most commercial carburizing is carried out in this temperature range.

Structure of Carburized Steel. — Carburizing steels ordinarily contain 0.10 to 0.20% carbon. At the temperatures indicated above the rates of carbon deposition and diffusion are such that the carbon content builds up to about 0.9 to 1.4% at the surface, and tapers off gradually or abruptly depending upon operating conditions and the characteristics of the steel, including the alloying elements present. As the carburizing temperature is increased, the rate of diffusion increases faster than the rate of carbon absorption, giving lower carbon contents at the surface and deeper penetration.

A slowly cooled carburized case is shown in Fig. 157. The black pearlitic area contains 0.8 to 0.9% carbon. The white network at the surface is Fe_3C, indicating a carbon content of perhaps 1.2% at

the surface.   The core retains its normal hypoeutectoid structure
of ferrite and pearlite.   (This photomicrograph does not extend to
the unaffected core structure.)

It should be noted that at the carburizing temperature of 1700° F.
the entire piece had a simple austenitic grain structure, the carbu-
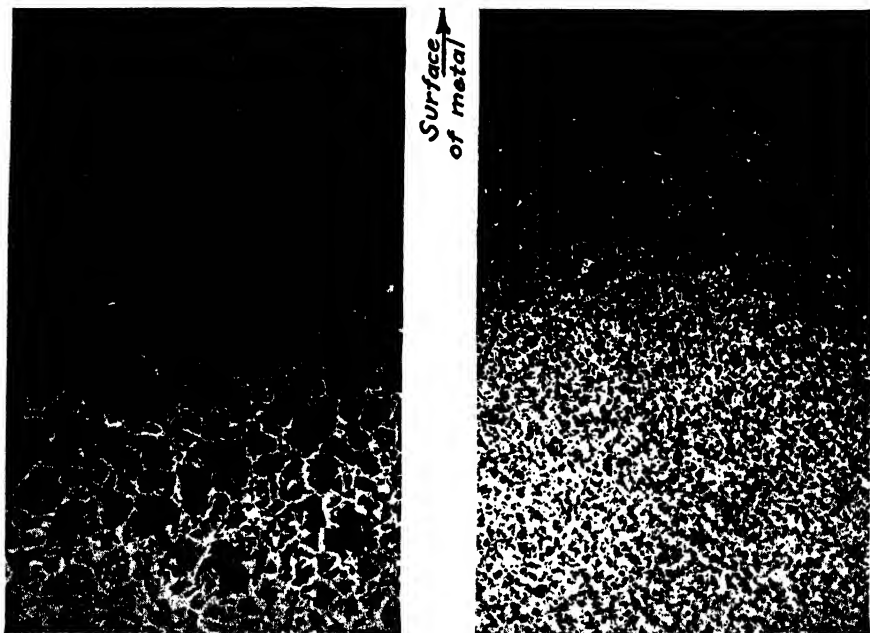
Left.   FIG. 157. — S.A.E. 1020, CARBURIZED AT 1700° F. FOR 8 HOURS, COOLED
SLOWLY IN THE BOX.   (Etched with picral.)   X50.

Right.   FIG. 158. — S.A.E. 2315, CARBURIZED AT 1650° F. FOR 5 HOURS, COOLED
SLOWLY IN THE BOX.   (Etched with picral.)   X50.

rized grains near the surface containing nearly their saturation limit
of carbon.   (See the Fe-Fe_3C diagram on page 158.)   Upon cooling
slowly, each portion of the specimen followed the reactions indi-
cated by the equilibrium diagram for a steel having an identical
carbon content; consequently, the normal or slowly cooled structures
of steels of all compositions between about 0.20% and 1.2%
carbon are represented.

A medium depth case (approximately 0.025 inch) without a
hypereutectoid layer is shown in Fig. 158.

While it is often desirable to cool slowly after carburizing and produce the relatively soft pearlitic structures illustrated above, carburized steels are always hardened by quenching before putting in service. The slow cooling operation is frequently omitted, whereupon hardening is effected by quenching directly from the carburizing heat. After quenching the appearance of the microstructure of a carburized section is greatly altered in that the high carbon martensitic region at the surface, being the most resistant to etching, is lighter in color than the core. (See Fig. 163, page 346.) Rather deep etching and high magnification generally reveal the hardened case as a typical martensitic structure with needlelike markings, although untransformed austenite may also be present.

The Carburizing Process. — Before considering the structure and heat treatment of carburized parts in more detail, certain features of the process itself will be reviewed.

Solid carburizing compounds usually have wood charcoal as a base and may contain bituminous coal, coke, and certain inorganic salts such as barium, calcium, and sodium carbonates. These salts act as energizers which speed up the process.

Davis [24] gives the following typical specification for a modern carburizing compound. (Davis also reviews the economic aspects of several carburizing processes.)

53–55% Hardwood charcoal
30–32% Solvay coke
2–3% Sodium carbonate
10–12% Barium carbonate
3–4% Calcium carbonate

The coke is essentially a filler or diluent but tends to reduce shrinkage of the compound and to improve its thermal conductivity. Another class of compounds contains higher charcoal contents without coke.

The most important function of the compound is to maintain a supply of CO gas at the surface of the steel. In pack carburizing practice the compound and the steel are generally packed together in alternate layers in a closed container. (Gas carburizing practice is considered later in this chapter.)

Loss of compound up to 10% may be expected due to burning, the greatest loss occurring when the direct quenching procedure is used. The tightness of the cover on the carburizing container is an important factor in controlling this loss. Other losses of a mechanical nature occur in handling, so that approximately 10 to 20% new compound is required for each heat under ordinary conditions.[25, 26]

Energizers. — The CO_2 liberated by the carburizing reactions given above is reduced to active CO gas by the red hot carbon present. The simplest explanation of the action of carbonates as energizers is that $BaCO_3$ and similar salts decompose to CO_2 and an oxide such as BaO, thus increasing the amount of CO_2 gas available for reduction to CO. Two other possibilities are the direct formation of CO according to reactions of the type:

$$Na_2CO_3 + C \rightleftharpoons Na_2O + 2CO,$$
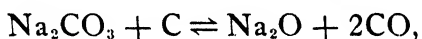
and the possible catalytic action of metal oxides such as Na_2O and BaO upon the reduction of CO_2 to CO by the carbon. The role of energizers is still a controversial subject.[20, 21, 27, 28]

Energizers may be made to adhere to the surface of the charcoal by the use of suitable binders such as tar and molasses. In another type of compound the ingredients are mixed in a pulverized state and used in the form of small pellets or " pills." In either case the compound particles are about the size of a pea.

Containers. — Carburizing containers may be iron or steel castings or welded sheet steel. Ordinary steel and iron containers become embrittled after prolonged exposure to carburization and are subject to warping and cracking as a result of continued heating and cooling through the structural transformation temperature range. It is often economical to use containers made of certain alloys of high nickel and chromium content which resist carburizing and warping. Such alloys range from the well known 18% chromium-8% nickel type, modified by the addition of silicon, to one containing 65% nickel-16.5% chromium. The life of these alloy containers must be several times that of unalloyed steels and irons in order to justify their high cost. For a complete discussion of carburizing containers see Roush and Dames.[29]

Selective Carburization. — Preparation of the parts to be car-
burized usually includes machining and a preliminary heat treatment
to relieve stresses and thereby reduce the distortion likely to occur in
carburizing.   If certain portions of the stock are to be left uncarbu-
rized, they may be protected by copper plating or by packing with
fireclay, sand, or other materials.   When copper plating is employed,
the entire piece may be plated and the portions to be carburized
given a final machine cut.   Alternatively, protective paints may be
applied before copper plating.   These paints then burn off and per-
mit carburizing to proceed on the portions not copper plated.   If
the cycle includes slow cooling from the carburizing heat, the cop-
per plating may be omitted provided sufficient stock is left to ma-
chine off the case in those portions which do not require harden-
ing.[30]

Typical Processing Schedule. — In selecting steels for machine
parts it is often possible to choose between carburizing and direct
hardening grades for a given application.   The operations neces-
sary to make a certain lever in a steering gear assembly were given
on page 302.   A corresponding lever for a light passenger car has
been made from S.A.E. 4620 steel purchased with a McQuaid-Ehn
grain size requirement of No. 5 to 7.   In this case the cam follower
was a stud forged integral with the lever, and the high hardness re-
quired for this rubbing contact surface was obtained by carburizing
according to the following schedule:

(1) Forge the blank.
(2) Heat with the furnace to 1750°–1800° F., hold at heat for
     ½ hour, cool in air.
(3) Tumble to remove scale.
(4) Rough machine.
(5) Remove grease by a chemical cleaning process.
(6) Copper plate.
(7) Machine all surfaces to be carburized.
(8) Carburize in boxes using solid compound.   Heat to
     1500° F. in 2 hours, hold at 1650°–1700° F. for 4 hours,
     or until the depth of the hypereutectoid plus eutectoid
     case is 0.04–0.05 inch.

(9) Cool to 1450°–1500° F. in the carburizing box and then quench in oil. Clean.

(10) Temper at 350° F. in a circulating air type furnace. Hold at heat for 1 hour.

(11) Draw the threaded end at 1100° F. by bringing it up to heat rapidly in a lead bath and quenching in water before the wearing surfaces are softened.

(12) Grit blast.

(13) Test every forging for hardness using a power driven Brinell type tester. The impression through the combined case and core must be less than 3.0 mm. in diameter.

(14) Test all wearing surfaces with a file. (They will be 58 to 62 Rockwell " C ".)

(15) Rechase threads, finish machine certain uncarburized parts and finish grind the hard surfaces.

Heat Treatment after Carburizing. — The high normalizing temperature in the above treatment is desirable, not only for relieving forging stresses, but to improve machinability by coarsening the grain structure. Reheating through the critical range in the carburizing heat will refine the grain size of this steel, even though the carburizing temperature is rather high. It should be noted that no grain refinement can be expected in the hardening operation itself since it does not, in this case, involve *heating* through the critical temperature range. It was found, however, that dropping the temperature to 1450°–1500° F. before quenching gave higher toughness and only slightly less combined case and core hardness compared with quenching directly from the box.

If the above process were applied to an S.A.E. 1020 steel of McQuaid-Ehn grain size No. 2 to 4, the part would probably be subject to operating failures because both the case and the core would be excessively coarse grained and brittle. Considerable refinement of the grain size of the case could be effected by lowering the normalizing temperature of the 1020 steel to 1650° F. and changing step (9) to:

(9a) Cool slowly in the box.

(9b) Reheat to 1450° F. and quench in water.

This would give a soft core and a finer grained case; however, if the carbon content of the case is well over the eutectoid composition there is the possibility that excess carbides will separate at grain boundaries, causing brittleness of the case and counteracting the normal tendency towards grain refinement and toughening upon reheating over the lower critical temperature.

Greater core hardness could be obtained by quenching directly from the box, followed by reheating and quenching from 1450° F. as above. However, if exceptionally good mechanical properties, including toughness, are required in both the core and case, the following treatments may be applied:

(9a)   Cool slowly in the box.
(9b)   Reheat to 1650° F. and quench in water.
(9c)   Reheat to 1450° F. and quench in water.

This is the well known double heat treatment for optimum mechanical properties in carburized steels. Both the core and the case are refined and hardened. In addition to refining the grain size of the core, treatment (9b) dissolves the excess carbides which may be present in the case and prevents their separation as cementite in the grain boundaries. (See also Chapter XV, page 523.) The disadvantages of the double heat treatment are high cost and the possibility of distortion upon heating and cooling through the critical temperature range.

When fine grained (McQuaid-Ehn test) steels are used, it is often desirable to use a single quench from the box, or from above the Ac_3 transformation of the core. This treatment gives high core hardness and strength and the toughness will be satisfactory provided the steel does not coarsen in the carburizing treatment. In fact, the toughness of the core may be greater than after treatments involving a second quench from *within* its critical temperature range (between Ac_1 and Ac_3 of the core). At 1450°–1500° F., the temperature range ordinarily used to refine the case in the double heat treatment, the low carbon core consists of ferrite plus high carbon austenite. Upon quenching from this temperature range, the austenitic regions of the core become martensitic while the soft ferritic regions remain essentially unchanged. The duplex hard-soft core

structure which results may be deficient in toughness. This is especially true when a single low temperature quench is used. Rapid or superficial heating for the low temperature quench may avoid serious changes in the core structure and improve the results obtained by heat treatments which involve a final quench to refine and harden the case.[31]

Although the single high temperature quench has several advantages when applied to fine grained (McQuaid-Ehn test) steels, there is the possibility of austenite retention which may cause low case hardness compared with the results obtained by using final quenching temperatures just above Ac_1. This condition is most likely to occur in alloy carburizing grades having hypereutectoid cases. It will be noted that at 1650°–1700° F. a carburized surface containing 1.2% carbon will consist of a homogeneous austenitic structure. Upon drastic quenching, a considerable proportion of austenite may be retained, thereby reducing the hardness. Cooling to 1400°–1450° F. before quenching will allow separation of excess cementite until the carbon content is reduced to the eutectoid composition, possibly 0.7% carbon for certain alloy steels. This condition is less conducive to austenite retention upon quenching. (See Chapter VIII. Note also that this is the treatment actually applied in the above processing schedule.)

It is evident that still other combinations are possible using ordinary quenching media, and, in addition, heat treatments involving hot-quenching have been studied.[32] Extended experimentation is often carried out to determine the best procedure for a given application. The problems of selecting a carburizing steel and determining the proper case depth and heat treatment after carburizing are considered further by Williams.[33]

Properties of Carburized Steels. — Typical mechanical properties of several S.A.E. carburizing steels are summarized in Table XLV. It may be noted that with the exception of S.A.E. 2515 the cases are all over 60 R " C " as quenched and after tempering at low temperatures, and that the principal differences in the steels are in core properties.

The grain size characteristics and alloy contents naturally influence the properties of both the core and case, and in general the

discussions in the preceding chapters apply to these steels.   In fact, the relationships between fine grain size in the McQuaid-Ehn test, structural abnormality, toughness, and inferior hardening characteristics were first observed in carburizing steels.   Fine grain size and the occurrence of structural abnormality in the McQuaid-Ehn test is not an indication of hardening difficulties in many of the alloy steels because their critical cooling rates are inherently low.

Although alloy carburizing steels have many natural advantages over carbon steels, the latter are extensively used for carburized parts wherever processing conditions and service requirements permit.   The free-machining grades are particularly popular.   Mc-Mullan [34] has summarized the characteristics of these steels as follows:

" Carbon steels require water or other drastic cooling media to produce maximum case hardness; the core of larger sections will not quench to a uniform structure and soft spots may occur in the case; abnormality even in small sections causes very soft spots in the case; the water hardening required produces distortion in thin sections and intricate shapes; carbon steels are not as strong and are more brittle than most alloy steels.   Steels X–1015 and X–1020 have a higher manganese range and are used for larger sections or wherever their higher hardenability is required.   Small parts are sometimes oil-quenched if case hardness requirements are not too high.   The higher manganese also improves machinability in soft low carbon steels.   Sulphur may be added purposely to increase machinability in which case manganese must be on the high side.   Steel made by the Bessemer process is quite apt to be abnormal and have soft spots in the hardened case but may be suitable for some applications.   Carbon steel in grades suitable for the application are used for all purposes where the superior or special properties of alloy steels are not required.

" Manganese steels tend to be coarse-grained and brittle but this difficulty can be mostly overcome by making the steel fine-grained by the ladle practice.   Manganese lowers the critical points making it possible to use lower quenching temperatures with the attending advantages.   These steels, like all the other alloy carburizing steels except those of medium carbon content, may be water quenched, but,

TABLE XLV. MECHANICAL PROPERTIES OF SEVERAL S.A.E. CARBURIZING STEELS HEAT TREATED AS 1 INCH ROUNDS

These data were taken from diagrams in " Bethlehem Alloy Steels " and are reproduced through the courtesy of Bethlehem Steel Company.

The steels were carburized at 1650°–1750° F., cooled in the box, reheated to the temperature indicated, quenched, and tempered as indicated in the first column. The tensile data and Brinell hardness refer to core properties while R " C " is the Rockwell hardness of the case.

T = tensile strength in 1000 lb./sq. in.
Y = yield point in 1000 lb./sq. in.
R = reduction in area, %
E = elongation in 2 inches, %

Temper ° F.	T	Y	R	E	Br	R " C "
S.A.E. 1015	Quenched in water from 1425° F.					
Quenched	96	63	50	19	201	67
250	96	63	50	20	201	65
350	95	62	50	21	197	63
500	91	55	55	22	183	60
S.A.E. 2315	Quenched in oil from 1450° F.					
Quenched	140	99	52	21	285	63
250	136	95	53	22	277	61
350	132	92	58	23	277	60
500	128	88	63	24	269	58
S.A.E. 2515	Quenched in oil from 1425° F.					
Quenched	169	155	51	13	363	58
250	169	154	52	13	363	57
350	168	153	53	13	352	55
500	165	148	55	18	341	53
S.A.E. 3115	Quenched in oil from 1425° F.					
Quenched	111	71	57	26	223	61
250	110	69	59	28	223	60
350	109	68	60	29	212	58
500	105	65	63	29	201	58
S.A.E. 4615	Quenched in oil from 1475° F.					
Quenched	119	91	59	22	248	64
250	118	88	60	22	248	62
350	118	86	61	22	242	60
500	116	79	63	24	229	58

TABLE XLV. (*Continued*)

Temper ° F.	T	Y	R	E	Br	R " C "
S.A.E. 6115	Quenched in oil from 1475° F.					
Quenched	103	61	58	28	207	62
250	103	61	58	28	207	60
350	103	61	58	28	207	59
500	101	59	62	30	201	57

unless the manganese content is too low or the section too large, may be oil-quenched also, lessening distortion as already mentioned. An outstanding advantage of manganese not obtainable from other alloying elements is its use in high sulphur free machining steels where its presence is necessary to prevent red shortness while the steel is being hot-worked. While all the manganese present may not be necessary for this purpose, better hot working properties are obtained than when manganese is low and better machinability may be obtained with low carbon contents."

Alloy Carburizing Steels. — The general effects of alloying elements on steels were discussed in Chapter IX and are summarized below from the standpoint of carburizing steels.

Nickel toughens and strengthens the steel. It lowers the hardening temperature which reduces the possibility of surface decarburization and of distortion in heat treatment. High nickel steels have a tendency to retain austenite in the case, especially upon quenching from high temperatures, thereby reducing the surface hardness. However, when properly heat treated, S.A.E. 2515 is suitable for highly stressed parts subject to heavy impact. (See Table XLV.)

Chromium increases the hardness and strength of the core and the wear resistance of the case. Chromium and the other carbide forming elements ordinarily increase the rate of carburization, while nickel reduces it somewhat. This factor is relatively unimportant compared with the characteristics of the products. The straight chromium steels were formerly considered too brittle for most carburizing applications; however, grain size characteristics may now be controlled by proper melting and ladle practice, and fine grained chromium steels of satisfactory toughness are attractive for carburizing because of their low cost, hardenability, and wear resistance.

The excellent combination of properties obtainable by adding

both nickel and chromium in proper proportions makes S.A.E. 3115 and the higher alloy types particularly suitable for carburizing.

Molybdenum has been used extensively with nickel in carburizing steels. Like chromium it adds strength, hardenability, and wear resistance to the toughness and other desirable features of nickel steels. Heavy duty automotive gears and bearings are frequently made of S.A.E. 4615 and 4815. Chromium-molybdenum steels have quite recently been introduced in carburizing grades.

Vanadium is used to refine the grain size, harden, and toughen the chromium-vanadium carburizing steels. The relatively high heat treating temperature required for optimum mechanical properties is a disadvantage from the standpoint of scaling, decarburization, and distortion. However, when hardened from cyanide baths, very tough wear resisting parts have been made of S.A.E. 6115.

McMullan [34, 35] has reported the properties of carburizing steels in considerable detail.

Gas Carburizing. — Carburizing by means of city gas, natural gas, propane, butane, and specially prepared gas mixtures is commercially successful and is gaining favor over solid pack methods.[36, 37, 38, 39] Both continuous [40, 41] and batch processes are in use.

A popular non-continuous or batch type of furnace consists of a cylindrical drum or retort of non-carburizing alloy steel which is rotated mechanically within an outer heat insulated cylinder. Heat is applied by means of gas flames in the annular space between the two cylinders. The carburizing gas passes through the sealed inner cylinder, where it reacts with the steel and is burned upon leaving the furnace. The furnace may be tilted for charging and removing the work. The mechanical movement of the work tends to keep the surfaces clean and to maintain uniform temperature and concentration of gases. (Solid carburizing compound may be used in the same type of furnace.)

A successful stationary batch type of furnace consists of an electrically heated vertical retort having a gas tight cover. The work is introduced into the furnace on racks or in containers. The carburizing gas may be circulated by a fan. In one type, oil is introduced into the heated retort where it is vaporized and circulated through the work as the carburizing medium.[42]

Many continuous gas carburizing furnaces are in service, especially in production lines where a continuous output of uniformly carburized parts is required. A notable example of a continuous unit capable of gas carburizing, cooling, reheating, quenching, and drawing has recently been described.[32]

Although gas carburizing is readily applied to small parts which are difficult to handle in large quantities when using solid compounds, it is also applicable to heavy parts. It has the advantage of rapid heating and offers the possibility of close control of the depth, carbon content, and carbon gradation of the case. The gases used are high in hydrocarbons rather than carbon monoxide, and they must be carefully compounded and uniform in composition for good carburizing characteristics. Excessive deposition of soot on the work by the heavy hydrocarbon gases has been a major problem. Dilution with cracked gas and burned gas products is one means of reducing soot.

Data on recent commercial practice have been assembled by Gable and Rowland [38] and are given in Table XLVI.

The total penetrations given in Table XLVI include the hypoeutectoid zone down to the point where the carbon content is approximately 0.4 to 0.5%, therefore the figures are somewhat higher than would be reported on the basis of hypereutectoid plus eutectoid depth. The temperatures used were 1650°–1750° F. The diffusion period represents time at heat after active carburization had been stopped. Its purpose is to increase the depth of penetration and reduce the carbon concentration at the surface. When all the variables involved in gas carburizing are under control, it is possible to obtain any desired case depth and carbon content at the surface.

Cyaniding. — Most liquid hardening baths have cyanide salts as their base. Their extremely poisonous nature must always be kept in mind when handling these materials. Hoods should be provided to carry off fumes. Parts to be hardened should be clean and dry since there is danger of explosive spattering if moisture is introduced with the work or tongs.

Low and medium carbon steels may be given a shallow but very hard case by heating from 15 minutes to 1 hour at temperatures

TABLE XLVI.   DATA ON COMMERCIAL GAS CARBURIZING INSTALLATIONS

Type of Furnace	Part Carburized	Gas Used	Type Steel	Case Depth Total Penetration	Total Time Cycle	Diffusion Period
Box-type Batch	Side Gears	Natural + DX	3115	0.045"	6 hrs.	none
Box-type Batch	Bearing Races	Natural + DX	4615	0.080"	10 hrs.	none
Box-type Batch	Piston Pins	Natural + DX	1020	0.045"	5½ hrs.	none
Box-type Batch	Clutch Parts	Raw Nat. & Nat. + DX	4620	0.030"	3 hrs.	none
Box-type Batch	Camshafts	Propane + CG	1020	0.070"	8½ hrs.	none
Horizontal Rotary	Bearing Parts	Raw Natural	4620	0.040" to 0.080"	4½ to 15 hrs.	1½ to 5 hrs.
Horizontal Rotary	Pins	Raw Natural	1020	0.040" to 0.125"	4 to 25 hrs.	none
Horizontal Rotary	Bearing Parts	Propane + CG	4620	0.050" to 0.120"	5½ to 19 hrs.	none
Horizontal Rotary	No Data	Raw Butane	No data	0.035"	4 hrs.	none
Stationary Vertical	Bearing Parts	Raw Natural	Krupp	0.070" to 0.210"	15 to 105 hrs.	5 to 35 hrs.
Stationary Vertical	Bearing Parts	Propane + CG	4620	0.100" to 0.150"	27 to 37 hrs.	2 hrs.
Stationary Vertical	Shafts	Raw Natural	1020	0.060"	7 hrs.	2 hrs.
Continuous	Plowshares	Butane + CG	1020	0.080"	10 hrs.	none
Continuous	Bearing Races	Propane + CG	4615	0.070"	10 hrs.	none
Continuous	Ring Gears	Natural + DX	4615	0.060"	7½ hrs.	none
Continuous	Camshafts	Propane + CG	4620	0.060"	7½ hrs.	none
Continuous	Ring Gears	Natural + CG	4615	0.045"	6 hrs.	none

DX is a flue gas prepared by combustion of natural gas under controlled conditions in a DX unit.
CG is a cracked gas prepared by external heating of a mixture of natural gas and air.

Courtesy of L. D. Gable and E. S. Rowland, Trans. Am. Soc. for Metals.[38]

from 1450°–1600° F., followed by quenching in water or oil. The salt bath ordinarily contains about 30% NaCN, the balance being NaCl and Na_2CO_3 which are inert as far as the hardening process is concerned. It is maintained at the proper concentration of cyanide by the addition of salts of high cyanide content, usually 96 to 98% NaCN.[43] The salt bath tends to give uniform heating.

The reactions occurring in cyanide baths are not known with certainty. The following steps indicating oxidation to sodium cyanate followed by decomposition to CO and N, the active agents, are given by Schlumpf.[22]

$$2NaCN + O_2 = 2NaCNO$$
$$4NaCNO + \text{heat} = 2NaCN + Na_2CO_3 + CO + 2N$$

Both carbon (from CO) and nitrogen are introduced into the steel from cyanide baths. A typical case will contain approximately 0.60% carbon and 0.50% nitrogen.[44] The nitrogen in cyanide cases is present as finely dispersed iron nitrides which impart high hardness and brittleness. The lower bath temperatures favor absorption of nitrogen but reduce penetration. Case depths of only 0.002 inch are sometimes used on medium carbon gears, and cyanided cases are seldom produced over 0.010 inch.

Bolts, washers, small gears, and other small parts are sometimes hardened superficially by applying powdered sodium cyanide while the steel is at a red heat. The case is likely to be non-uniform and modern liquid baths are preferred.

The speed, low cost, high superficial hardness, and cleanliness of the hardened surface make the cyanide process readily adaptable to production heat treatment when-
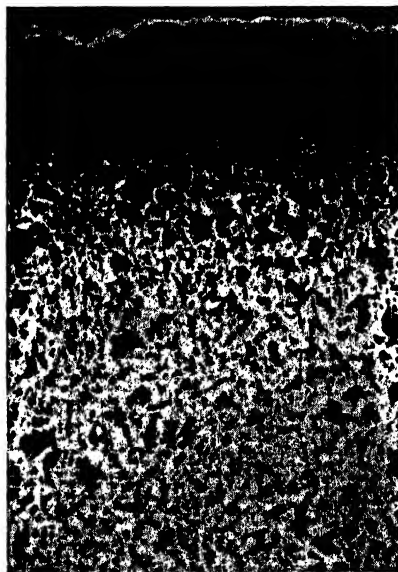
FIG. 159. — S.A.E. 1020, CARBU-RIZED, SHOWS SLIGHT SURFACE DECARBURIZATION. (Etched with picral.) X40.

ever the service requirements of the part permit the use of a very thin case.

Cyanide baths are often used for heat treatment of medium carbon and even carburized steels to prevent decarburization and to take advantage of the clean surface and high skin hardness obtained after quenching. An example of skin decarburization of a pack carburized steel is illustrated in Fig. 159. The white ferrite band at

Left. FIG. 160. — S.A.E. 1115, CYANIDED CASE. (Etched with picral.) X100.
Right. FIG. 161. — SAME AS FIG. 160, REHEATED TO 1650° F., AND COOLED SLOWLY. SHOWS NITRIDE NEEDLES NEAR THE SURFACE. (Etched with picral.) X500.

the surface was caused by infiltration of excess air into the carburizing box through a poorly sealed cover. (This piece was not finish machined before carburizing and the soft layer will be removed in the final grinding operation after hardening.) A similar surface condition may be produced when carburized parts are reheated for hardening without proper atmospheric control or surface protection. Both scaling (oxidation of the surface) and decarburization [45] must be considered when adjusting the atmospheres of furnaces for heat treating carburized parts. In the case of highly

finished gears, for example, a soft skin is particularly undesirable, therefore heating in cyanide is widely practiced.

A typical cyanide treatment for a small part made of S.A.E. 1115 steel consists of heating for 30 minutes in a 30% NaCN salt bath held at 1500°–1550° F., followed by quenching in water. The resulting case, shown in Fig. 160, is 0.004 inch thick and file hard. The core structure is banded in the direction of rolling as indicated by the slag stringers and alternate ferrite and martensite areas. Quenching from above instead of from within the critical temperature range would have resulted in a more homogeneous core structure.

The above steel was reheated to 1650° F. and cooled slowly. The presence of nitrogen in the coating is indicated in Fig. 161 by the iron nitride needles which have been coarsened considerably by this treatment. A single band of pearlite is also visible in the photomicrograph.

A cyanide case on a low carbon steel stamping is shown in Fig. 162.

FIG. 162. — LOW CARBON STEEL SHEET WITH CYANIDED CASE. (Etched with 3% nital.) X100. (*Courtesy of W. L. Havekotte.*)

Activated Cyanide Hardening Baths. — Recent improvements in carburizing include the use of several new liquid hardening baths. These are usually based on molten cyanide salts to which activating agents are added. According to Beckwith [46] these activating agents are alkaline earth metals combined with the halogens or cyanogen. The compositions of the commercial salt baths vary considerably and the reactions in activated baths are more numerous and less definitely known than the simple cyaniding reactions. In addition to sodium and potassium cyanides, the essential ingredients of various baths include calcium cyanide, calcium cyanamide, barium chloride, calcium and sodium chlorides, and polymerized hydrocyanic

TABLE XLVII. TYPICAL APPLICATIONS AND REQUIREMENTS OF
CASE HARDENED PARTS FOR PASSENGER AUTOMOBILES

Case Depth	Type of Steel	Typical Parts	Surface Condition When Treated	Requirements
Up to 0.020″	Low Carbon	Push Rod Balls. Push Rod Sockets. Speedometer Gears. Shackle Bolts. Shifter Forks. Tie Rod Ball Studs. Tie Rod Ball Stud Sockets.	Finished before treatment except for possible polishing operation.	High degree of wear resistance. Clean surface after treatment.
		Water Pump Shafts.	When nitrided 0.0003″ to 0.001″ per side removed for maximum wear resistance.	——
	Medium Carbon	Transmission Gears. Transmission Main Shafts and Synchronizer Sleeves.	Generally finished before treatment.	Maintain or increase surface carbon for higher degree of wear resistance and greater fatigue resistance. Resistance to high occasional or shock loading. Clean surface after treatment.
0.020″ to 0.040″	Low Carbon	Pitman Arm Ball Studs. Intermediate Steering Arm Ball Studs.	Finished before treatment except for possible polishing operation.	High degree of wear resistance. Sufficient strength to withstand service loads. Clean surface after treatment.
		Steering Arm Bushings. Valve Rocker Arms. Valve Rocker Arm Shafts. Gear Shifter Shafts. Brake and Clutch Pedal Shafts. Water Pump Shafts.	Generally semi-finished before treatment and finish ground after treatment.	Enough case to permit finish grinding and maintain high hardness for necessary wear resistance.

As service becomes more severe as in trucks and buses, the depth of case for corresponding parts may be somewhat deeper and the general tendency is to change to a steel which permits of better physical properties, especially in the core.

TABLE XLVII.   (*Continued*)

Case Depth	Type of Steel	Typical Parts	Surface Condition When Treated	Requirements
0.020″ to 0.040″	Medium Carbon	Transmission Gears. Ring Gears. Drive Pinions.	Finished before treatment except for lapping operation.	Increase carbon content at surface for greater wear, load carrying ability, and fatigue resistance. Clean surface after treatment.
0.040″ to 0.060″	Low Carbon	Ring Gears. Drive Pinions. Transmission Gears. Side Gears. Side Pinions.	Finished before treatment except for lapping operation.	High wear resistance to resist sliding, rolling, or friction of abrasive nature. Resistance to heavy crushing loads. Resistance to alternate bending stresses of high magnitude.
		Piston Pins. Steering Knuckle Pins.* Side Pinion Axles.* Reverse Idler Gear Shafts.* Counter Gear Shafts.*	Semi-finished before treatment. Finish ground after treatment.	Operate frequently in bronze bushings. High wear resistance. Resistance to relatively high unit loads that cause Brinelling. Resistance to fairly high alternate bending stresses. King pins also require resistance to high instantaneous or shock loads.
		Roller Bearings.	——	Resistance to heavy rolling action.
Over 0.060″	Low Carbon	Camshafts.**	Semi-finished before treatment.	High wear resistance. Resistance to shock. Resistance to relatively high unit loads with a rapid sliding action.

* More case than actually necessary to satisfy service requirements put on parts of such length as these in order to avoid necessity of straightening to extremely accurate degree before grinding.

** Carburize deeper than necessary to satisfy service requirements to compensate for inaccuracies of machining due to length of these parts, distortion due to heat treatment and insure sufficient case after grinding for necessary hardness and wear resistance.

Courtesy of A. L. Boegehold and C. J. Tobin, Trans. Am. Soc. for Metals.[44]

acid.[22, 47]   According to reactions given by Schlumpf [22] for barium and sodium salts the final reactive agents are CO and N as in ordinary cyanide baths; however, the nitrogen content of the case is generally less than one-half that of cyanided cases.   Eutectoid and slightly higher than eutectoid carbon contents with approximately 0.25% nitrogen have been obtained.   The mechanical characteristics of activated liquid bath cases resemble those of carburized rather than cyanided cases.

The temperature range of approximately 1500°–1650° F. which is used for liquid baths is lower than the usual carburizing temperature range, and the time is generally short — less than two hours.   Although activated baths are used mainly for thin case work, it is possible to obtain 0.020 to 0.040 inch case depths.

FIG. 163. — S.A.E. 1035, CHAPMANIZED. SHOWN IN THE HARDENED CONDITION.   (Etched with picral.) X50.   (Sample furnished by H. H. Lurie.)

The advantages of liquid bath hardening over pack carburizing are rapid heat transfer and freedom from oxidation during heating; rapid absorption of hardening agents at the steel surface; the hardening action of nitrogen in addition to the carbon; protection from oxidation or decarburization during transfer to the quenching medium by the film of cyanide which adheres to the surface; and the saving due to direct quenching. The lower temperatures used reduce distortion but also reduce the possible depth of penetration because of the lower diffusion rate.

The Chapmanizing process [48] differs from those mentioned above in that dissociated ammonia gas is bubbled through the molten salt bath which tends to increase the nitrogen content of the steel.   Although Chapmanizing is used for relatively thin cases, compared with carburizing, it is much faster than nitriding (see page 351) and

high hardnesses, approaching those obtained by nitriding, have been reported. A Chapmanized case produced on a 0.35% carbon steel is shown in Fig. 163.

Applications of Cases of Various Depths. — The engineer or metallurgist must consider steels of many different chemical compositions, an increasing number of carburizing processes, and a suitable sequence of operations before putting a case hardened part into production. Service conditions and design limitations must be given first consideration in choosing a workable procedure. Both of these factors influence the required depth of case. Boegehold and Tobin [44] have classified case depths as shown in Table XLVII.

Thin Cases. — In general, parts which require high surface hardness for wear resistance but are not subjected to high concentrated crushing loads may be hardened superficially. Small stampings, screws, gears, bushings, etc., are given shallow cases in order to preserve sufficient core thickness. Bolts and nuts may be case hardened in this manner to prevent scoring. The ordinary molten cyanide or the activated cyanide baths may be used for cases less than 0.010 inch deep. The nitriding process is also available for thin cases where exceptional hardness is needed. Gas carburizing is favored over pack methods for the production of thin carburized cases.

It will be noted in Table XLVII that medium carbon steels are given a thin or medium depth case in the production of certain automotive gears and shafting. While case hardening was formerly applied only to low carbon steels, there is an increasing tendency to surface harden medium carbon (0.25–0.45% carbon) steels in order to increase their wear resistance and to avoid the possibility of producing a thin decarburized skin during heat treatment.

Automobile transmission gears are an example of the latitude of present day metallurgical practice. The following combinations have all been used successfully:

0.15–0.25% carbon steel — Medium heavy case (0.040″–0.060″).

0.25–0.35% carbon steel — Medium case (0.020″–0.040″).

0.35–0.45% carbon steel — Thin case (up to 0.020″).

0.45–0.55% carbon steel — Thin cyanide case or hardened without surface treatment.

Medium Cases. — When it is desirable to finish grind after heat treatment in order to maintain close dimensional tolerances, a medium case is necessary to provide a sufficient depth of hardened metal for wear resistance after grinding. The resistance to crushing of medium depth cases will not be great unless a high strength core is provided. Both gas and pack carburizing are suitable for medium depths and excellent results are obtained with the activated cyanide baths.

Medium Heavy and Heavy Cases. — Case depths over 0.04 inch will withstand high crushing loads and will allow considerable grinding of distorted parts. Medium heavy cases may be applied by both gas and solid pack processes. The heavy cases required on large steel gears for mining machinery and other heavy duty applications are produced by pack carburizing methods.

Nitriding. — It will be remembered that in certain carburizing and cyaniding processes nitrogen is introduced into the steel in addition to the carbon. In the nitriding process nitrogen is responsible for the entire hardening effect.

Much of the early work on nitriding, as carried out in this country, was reported in the " Nitriding Symposium " of 1929 [40], which indicates the recent origin of this important metallurgical process.

Nitriding has several advantages over carburizing as a hard surfacing process, including greater hardness and wear resistance, bet-

TABLE XLVIII.   COMPOSITIONS AND MECHANICAL PROPERTIES
OF COMMONLY USED NITRIDING STEELS

	N125 (Type H)	N125N [a] (Type H with Nickel)	N135 (Type G)	N135 Modified [a]	N230
Carbon	0.20–0.30	0.20–0.27	0.30–0.40	0.38–0.45	0.25–0.35
Manganese	0.40–0.60	0.40–0.70	0.40–0.60	0.40–0.70	0.40–0.60
Silicon	0.20–0.30	—	0.20–0.30	—	0.20–0.30
Aluminum	0.90–1.40	1.10–1.40	0.90–1.40	0.95–1.35	1.00–1.50
Chromium	0.90–1.40	1.00–1.30	0.90–1.40	1.40–1.80	—
Molybdenum ..	0.15–0.25	0.20–0.30	0.15–0.25	0.30–0.45	0.60–1.00
Nickel	—	3.25–3.75	—	—	—

[a] Aircraft specifications.

Letter " N " preceding a number indicates nitralloy steels.

TABLE XLVIII. (*Continued*)

PROPERTIES OF N125, N135, AND N135 MODIFIED

Tempering Temp., ° F.	Yield Point, lb./sq. in.	Tensile Strength, lb./sq. in.	Elong. in 2 in., %	Reduction of Area, %	Charpy Impact, ft.-lb.	Brinell Hardness No.
N125 [a]						
800	156 300	178 000	12	47	20.6	400
900	143 500	170 000	16	49	25.5	365
1000	133 300	160 000	17	54	30.1	340
1100	117 500	134 500	18	61	48.1	290
1200	103 800	122 000	21.5	67	59.2	255
1300	85 500	102 500	27	72	69.0	228
1400	75 000	90 500	32	73	73.0	190
Annealed	60 000	80 000	34	70.5	47.8	157
N135 [b]						
800	180 000	224 500	11	36	12	445
900	165 000	206 300	11.5	37.5	15	415
1000	158 000	182 500	15	50	22	363
1100	137 500	156 000	16.5	57	35.2	330
1200	120 000	138 000	20	60	44.0	285
1300	103 300	121 000	23	62.5	55.3	226
1400	80 500	104 300	28	59	54.5	200
Annealed at 1450	69 000	95 000	30	67.5	31.6	186
N135 Modified [c]					Rockwell "C" Hardness	
1000	160 000	185 000	12.5	42.0	40	375
1100	133 000	157 000	16.5	50.5	33–34	321
1200	115 000	132 500	19.5	56.5	27–28	286
1300	84 000	109 000	25.5	60.0	20	228

[a] N125 — 0.23% C, 0.51% Mn, 0.20% Si, 1.58% Cr, 1.24% Al, 0.20% Mo, 0.011% P, 0.011% S. Quenched in oil from 1750° F. Held at temperature 30 minutes.

[b] N135 — 0.36% C, 0.51% Mn, 0.27% Si, 1.49% Cr, 1.23% Al, 0.18% Mo, 0.010% S, 0.013% P. Quenched in oil from 1650° F.

[c] N135 Modified — Quenched in oil from 1700° F.

Courtesy of V. O. Homerberg, Metals Handbook.[50]

ter corrosion resistance, better retention of hardness at elevated temperatures, and less distortion in the hardening process. On the other hand, there are certain factors which increase the costs of the process and limit its application. For example, the simple carbon

steels which are often used for carburizing are not suitable for nitriding. Furthermore, the practical limit for case depth is about 0.035 inch, which requires a 90 hour treatment. Both the process and the steels used are patented.

Nitriding Steels. — The compositions and mechanical properties of the most commonly used nitriding steels are given by Homerberg [50] and are reproduced in Table XLVIII.

In order that they may withstand high crushing loads it is necessary that nitriding steels have high core hardnesses because their cases are ordinarily quite thin. To obtain a satisfactory core hardness after tempering at the high temperatures required, nitriding steels are made with higher carbon contents than carburizing steels.

It is apparent that aluminum is a major alloying element in nitriding steels. In the earliest investigations of nitriding it was found that aluminum had the strongest tendency to form nitrides of any of the alloying elements available, and recent work has substantiated the fact that the precipitation of the compound AlN in a very finely divided state is largely responsible for the extreme hardness of the aluminum containing nitriding steels.[51]

The familiar carbide forming elements Cr, V, Mo, and W are also nitride formers and are used in nitriding steels. Certain of the tool and die steels, which ordinarily contain these elements, are being successfully nitrided for additional surface hardness.

Nickel tends to toughen the case and decrease its hardness slightly. It does not contribute to case hardness through nitride formation. The principal function of nickel in nitriding steels is hardening and strengthening of the core, thereby giving better support to the case. When nickel is added to the aluminum containing types, hardening and strengthening of the core occurs during the nitriding treatment due to the precipitation of a nickel-aluminum compound.

Sulphur and selenium may be added to nitriding steels to improve their machinability.

The Nitriding Process. — In contrast with the variety of gas mixtures available for gas carburizing, ammonia gas is almost universally used in the nitriding process. It is purchased in tanks as liquid ammonia and introduced into the furnace as a gas at slightly greater

than atmospheric pressure. At the furnace temperatures of 850°–
1200° F. the gas dissociates partially as follows:

$$2NH_3 \rightleftharpoons 2N + 3H_2$$
$$2N \rightleftharpoons N_2$$
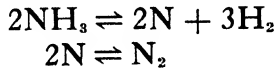
The dissociation of ammonia is ordinarily held to approximately
30% but may reach 95% in certain continuous furnaces. The
atomic nitrogen is responsible for surface hardening, therefore the
dissociation must take place in the nitriding chamber. It has been
found that at the higher temperatures, 1050° to 1200° F., the depth
of penetration is increased, while at temperatures below 1000° F.
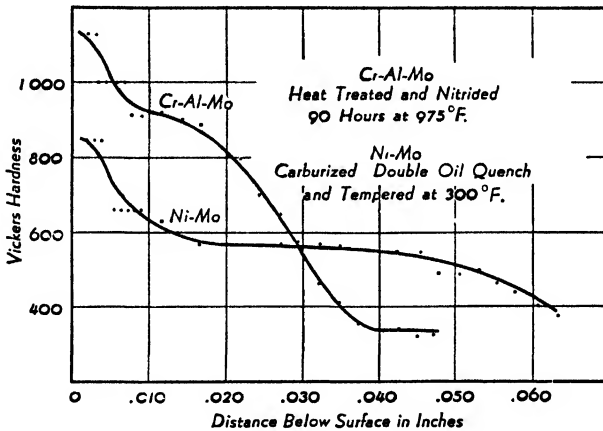
Fig. 164. — Comparison of the Hardness Penetration Characteristics of
Alloy Nitriding and Carburizing Steels. (*Courtesy of Robert Sergeson, Trans.
Am. Soc. for Steel Treating.*[52])

the maximum hardness is obtained. Various operating cycles have
been proposed in which a high temperature period is either pre-
ceded or followed by a low temperature period. In the continuous
furnace it is possible to increase the temperature gradually during
the entire treatment. The time required is much longer than for a
carburized case of the same depth, treatments up to 96 hours being
used in certain instances.

The hardness-depth characteristics of typical nitrided and car-
burized cases are given in Fig. 164.[52] These data are for high
quality alloy steels of suitable composition for the process used. Al-

though nitrided cases are characteristically thin, properly nitrided steels maintain their superior hardness to depths sufficiently great to permit some wear.

The following schedule of operations for processing pump shafts and other small parts of simple design is given in the material specifications of a prominent builder of airplane engines. The alloy used is N135 (Nitralloy G) supplied as annealed bars.

(1) Rough machine (if necessary) leaving at least ⅛″ stock on each surface.

(2) Heat to 1750° F., hold 2 hours per inch of cross section, quench in oil.

(3) Draw at 1050°–1075° F.

(4) Machine to within 0.000–0.003″ on each surface to be hardened. Leave $\frac{1}{16}$″ or more on surfaces which are to be left soft.

(5) Soak at 1025°–1050° F., 2 hours for each inch of section, to relieve strains.

(6) Polish surfaces to be hardened, to remove oxide film.

(7) Nitride at 950°–975° F., 40 to 60 hours.

(8) Grind off case from areas to be left soft, and finish machine.

The high hardening temperature is necessary to put the alloying elements present into solution in the austenite, giving good core toughness and strength after quenching and drawing. The drawing temperature of operation (3) is lower than that usually recommended; namely, 1200°–1250° F. The drawing or tempering heats before and after finish machining, operations (3) and (5), accomplish three purposes. First, the final core properties are established since the nitriding temperature is lower than the draw temperature and little further softening may be expected. Second, the sorbitic microstructure obtained is necessary for the production of a tough case. The free ferrite present in annealed structures results in brittle nitrided cases. It is especially important to avoid decarburization of the surface of nitriding steels for the same reason. Third, internal stresses set up during machining and hardening are relieved, reducing the subsequent distortion during nitriding to a minimum. In this respect the nitriding process is much superior to carburizing

in that practically no distortion occurs after finish machining.   A slight increase in size accompanies nitriding; however, the growth is constant for a given part and can usually be provided for in finish machining.

In operation (4) above provision was made for grinding off the case in areas to be left soft.   Alternative methods for preventing hardening of local areas include tin or nickel plating or coating with a thin layer of tin or solder.

The core hardness is given as over 30 R " C " before nitriding and over 28 R " C " after nitriding, corresponding to about 290 to 270 Brinell.   The case must be dead hard when tested with a sharp file. It will be noted that no heat treatment is required to produce hardness after nitriding.   The work is cooled either rapidly or slowly after operation (7).

Characteristics of Nitrided Steels. — Nitrided steel has several desirable properties in addition to surface hardness and wear resistance.   Its corrosion resistance is such that nitrided steel is often preferred over carburized steel for valves, pump parts, gears, shafts, and bearings operating in steam, crude oil, engine oil and gasolines, alkali, and combustion products.   It has been found that the best corrosion resistance is obtained when the parts are used as they come from the nitriding container without grinding or lapping.   On the other hand, maximum hardness and toughness of case are obtained if about 0.001 inch is removed.

Nitriding raises the endurance limit, especially when bending stresses are involved.   It has been reported that the effect of scratches, notches, small fillets, etc., on the fatigue strength is greatly reduced after a part has been nitrided, so that the endurance limit as ordinarily determined under ideal test conditions may be used with reasonable assurance even though service conditions involve surface stress concentrations.   Nitrided steels retain their fatigue strength under corrosive conditions to a greater extent than other case hardened steels.   Furthermore, the corrosion-fatigue resistance required in some of the above mentioned applications is combined with a wear resistance not readily developed in other well known alloys having inherently better corrosion resistance.

Many applications of nitrided steels depend on their stability and

resistance to softening after long heating at temperatures up to about 900° F. Hardness measured at elevated temperatures (hot-hardness) is also undiminished up to 750° F. and remains higher than that of high speed steel when tested at temperatures up to 950° F.[53] Carburized cases lose their hardness rapidly through tempering at elevated temperatures.

The limited application of nitriding compared with carburizing can be attributed mainly to the increased cost of the process and of the special alloy steels required. The furnace costs are relatively high due to the long time at temperature and to the need for expensive high alloy containers which resist nitriding. Porcelain enameled containers have also been used. Savings are sometimes possible, compared with carburizing practice, due to elimination of straightening and finish grinding or lapping after nitriding.

Some specific applications of nitrided parts include automobile and airplane piston pins, push rods, valves, valve stem guides, pumpshafts, camshafts, crankshafts, cylinder liners, and timing gears. Similar parts are nitrided for diesel engines, supplemented by several additional parts of the fuel injection mechanism.

Other applications include valve parts for high temperature, high pressure steam service and for the oil refining industry; also gages, mandrels, chucks, spindles, bushings, clutches, gears, die casting dies, wire drawing and extruding dies, forming dies, and ball and roller bearing parts.[50, 53]

REFERENCES

1. Burns, R. M., and Schuh, A. E., " Protective Coatings for Metals," Reinhold Publishing Corp., New York (1939)
2. Peirce, W. M., " Zinc Coatings," Metals Handbook, 1111, Am. Soc. for Metals, Cleveland (1939)
3. Thum, E. E., " Ductile Zinc Coating Deposited on Wire by Electrolysis," Metal Progress, 31, 524 (1937)
4. Heussner, C. E., " Protection of Base Metals by the Use of Metallic Coatings " — and — Wirshing, R. J., " The Pretreatment of Metals," Symposium on Protecting Metals Against Corrosion, Am. Soc. for Testing Mats., Philadelphia (1938)
5. Kelley, F. C., " Zinc Impregnation," Metals Handbook, 1087, Am. Soc. for Metals, Cleveland (1939)
6. Gonser, B. W., " Tinning of Steel," Metals Handbook, 1106, Am. Soc. for Metals, Cleveland (1939)
7. McKay, R. J., and Worthington, R., " Corrosion Resistance of Metals and Alloys," Reinhold Publishing Corp., New York (1936)

8. Sayles, B. J., "Aluminum Impregnation," Metals Handbook, 1082, Am. Soc. for Metals, Cleveland (1939)

9. Tieman, H. P., "Iron and Steel (A Pocket Encyclopedia)," McGraw-Hill Book Company, New York (1933)

10. Young, C. B. F., "Coloring Steel," Metals Handbook, 1118, Am. Soc. for Metals, Cleveland (1939)

11. Andrews, A. I., "Enamels," The Twin City Printing Co., Champaign, Ill. (1935)

12. Ihrig, H. K., "Silicon Impregnation of Iron and Steel," Metal Progress, **33**, 367 (1938) — and — "Silicon Impregnation," Metals Handbook, 1090, Am. Soc. for Metals, Cleveland (1939)

13. Sarbey, M. D., "Metal Coating by Cementation," Metal Progress, **34**, 447 (1938)

14. Kelley, F. C., "Chromium Impregnation," Metals Handbook, 1085, Am. Soc. for Metals, Cleveland (1939)

15. Schueler, J. L., "Galvanized Sheets and Galvannealed Sheets," Metal Progress, **31**, 499 (1937). See also **32**, 162 (1937) and **34**, 53 (1938)

16. Camp, J. M., and Francis, C. B., "The Making, Shaping and Treating of Steel," Carnegie Steel Company, Pittsburgh (1925)

17. Spencer, W. H., "Wear Resistance," Steel, **103**, 60 (Dec. 5, 1938)

18. Burns, R. M., "Sprayed Metal Coatings," Metals Handbook, 1115, Am. Soc. for Metals, Cleveland (1939)

19. "Carburizing Symposium," Trans. Am. Soc. for Metals, **26**, 427 to 759 (1938)

20. Grossmann, M. A., "A Review of Some Fundamentals of Carburizing," Trans. Am. Soc. for Metals, **26**, 427 (1938)

21. McQuaid, H. W., "A Theoretical Discussion of the Action of Solid Carburizing Agents," Trans. Am. Soc. for Metals, **26**, 443 (1938)

22. Schlumpf, R. W., "The Mechanism of Carburization," Metals Handbook, 1037, Am. Soc. for Metals, Cleveland (1939)

23. Grimshaw, L. C., "Diffusion of Carbon from Steel into Iron," Trans. Am. Inst. Mining and Met. Eng., **131**, 263 (1938)

24. Davis, E. F., "Production Carburizing," Trans. Am. Soc. for Metals, **26**, 607 (1938)

25. McQuaid, H. W., "Pack Carburizing," Metals Handbook, 1051, Am. Soc. for Metals, Cleveland (1939)

26. Widrig, S. L., "The Physical and Chemical Characteristics of Carburizing Compounds and their Handling in Production," Trans. Am. Soc. for Metals, **26**, 687 (1938)

27. Ragatz, R. A., and Kowalke, L. O., "Mechanism of Energizer Action in Carburization," Metals and Alloys, **2**, 343 (1931)

28. Enos, G. M., "The Role of Energizers in Carburizing Compounds," Trans. Am. Soc. for Steel Treating, **20**, 27 (1932)

29. Roush, R. W., and Dames, A. C., "A Study of Commercial Carburizing Containers," Trans. Am. Soc. for Metals, **26**, 646 (1938)

30. "Selective Carburization of Steels," Metals Handbook, 1068, Am. Soc. for Metals, Cleveland (1939)

31. Bleakney, H. H., "Brittleness of Low Carbon Steel," Trans. Am. Soc. for Steel Treating, **20**, 321 (1932)

32. Queneau, B. R., and Dowdell, R. L., "Case Hardening by Direct Transformation of Austenite," Trans. Am. Soc. for Metals, **25**, 904 (1937)

33. Williams, G. T., " Selection of Carburizing Steel, Case Depth, and Heat Treatment," Metals Handbook, 1043, Am. Soc. for Metals, Cleveland (1939)

34. McMullan, O. W., " Steels Used in the Carburizing Process," Trans. Am. Soc. for Metals, **26**, 546 (1938)

35. McMullan, O. W., " Physical Properties of Case Hardened Steels," Trans. Am. Soc. for Metals, **23**, 319 (1935). See also **24**, 262 (1936)

36. McQuaid, H. W., " Gas Carburizing," Metal Progress, **24**, 22 (July 1934)

37. Dow, J. A., " Gas Carburizing," Metals Handbook, 1054, Am. Soc. for Metals, Cleveland (1939)

38. Gable, L. D., and Rowland, E. S., " Commercial Gas Carburizing," Trans. Am. Soc. for Metals, **26**, 706 (1938)

39. Williams, G. T., " Gaseous Media for Carburizing," Trans. Am. Soc. for Metals, **26**, 463 (1938)

40. Cowan, R. J., " Continuous Carburizing by Gas," Metal Progress, **21**, 44 (Feb. 1932) -- and -- " Notes on Continuous Gas Carburizing," Trans. Am. Soc. for Metals, **25**, 843 (1937)

41. Sanders, W. E., Coyle, J. L., and Darrah, W. A., " A Practical Muffleless Carburizing Furnace," Metal Progress, **34**, 678 (1938)

42. Lindberg, A. J., " Oil as a Carburizing Medium," Metals and Alloys, **2**, 106 (1931)

43. Northrup, H. B., " Cyaniding," Metals Handbook, 1063, Am. Soc. for Metals, Cleveland (1939)

44. Boegehold, A. L., and Tobin, C. J., " Factors Governing Selection of Type of Carburized Case," Trans. Am. Soc. for Metals, **26**, 493 (1938)

45. Webber, J. A., " Furnace Atmospheres and Decarburization," Trans. Am. Soc. for Metals, **26**, 515 (1938)

46. Beckwith, B. B., " Liquid Bath Carburizing," Trans. Am. Soc. for Metals, **26**, 752 (1938)

47. Davis, E. F., " Liquid Carburizing," Metals Handbook, 1060, Am. Soc. for Metals, Cleveland (1939)

48. Rolf, R. L., " Chapmanizing," Steel, **102**, 48 (Jan. 10, 1938)

49. " Nitriding Symposium," Am. Soc. for Steel Treating, Cleveland (1929)

50. Homerberg, V. O., " Nitriding," Metals Handbook, 1071, Am. Soc. for Metals, Cleveland (1939)

51. Norton, J. T., " X-ray Study of the Action of Aluminum during Nitride Hardening," Trans. Am. Inst. Mining and Met. Eng., **113**, 262 (1934)

52. Sergeson, R., " Investigations in Nitriding," Trans. Am. Soc. for Steel Treating, **16**, 145 (1929)

53. Harder, O. E., " Nitriding for the Engineer," Metals and Alloys, **2**, 132 (1931)

CHAPTER XI

STEEL CASTINGS

Cast steel structures occur in three widely different products of the steel industry; namely, ingots, steel castings, and fusion welds.

Steel ingots, from which all rolled and forged steel products are made, are cast in metal ingot molds. The design of these molds is based mainly on the metallurgical and processing requirements of the metal, since the physical shape of the product ordinarily bears little resemblance to the shape of the ingot. The shrinkage cavity which forms at the top of killed and semi-killed steel ingots is removed by cropping during rolling, and the microscopic cavities which may form at the interstices of the dendrites are usually closed in the hot working operations. The subject of ingot making and ingot structure, which is of prime importance to steel makers, has been omitted from this text along with many other interesting problems in production metallurgy which are closely related to the quality of metallurgical products.

At the other extreme from the standpoint of average size are fusion welds, which have been used to join practically all types of metal products. Some of the metallurgical features and applications of welding will be considered in the next chapter.

Steel castings, the product of the steel foundry, are of more or less intricate shape and are cast in sand molds. The steel is usually melted in open hearth or electric furnaces. (The crucible and converter processes were used in the early steel foundries.) Although the conditions under which such castings solidify are, in certain respects, not as favorable as in ingot making, steel castings of excellent quality may be made by a combination of proper design, good foundry practice, and careful experimentation with individual castings.

Steel castings range in size from very small articles to structures over 150 tons in weight, requiring the output of several open hearth furnaces of the size ordinarily used in steel foundries.

General Application of Steel Castings. — One ordinarily thinks of the automobile industry as the largest user of steel bars, forgings, and stamping sheets.   In like manner steel castings are associated with the railway industry, and particularly with locomotives in which they are used for frames, driving wheels, cross heads, and many other smaller parts.   Data by Rogers[1] show that in 1929 about 40% of the commercial steel castings produced in the United States were used by the railroads and about 10% were used for rolling mills.   Less than 5% were used for each of the following purposes: buckets, shovels, and dredges; tractors and trailers; cranes, hoists, and derricks; refinery and oil-well equipment; valves and fittings; automotive equipment; electrical machinery; roadmaking machinery; presses; machine tools; and agricultural machinery.   This classification does not include castings made in foundries of automobile, machine tool, and other manufacturing plants for their own use.

The total production of steel castings of all kinds has been reported as 1,583,000 gross tons in 1929, including 192,900 tons, or 12.2%, of alloy cast steels.[2]   Data on alloy cast steel production have not been readily available since 1933, but the total production for 1937 has been estimated as 1,027,800 gross tons, 15.5% being alloy grades.[2]   In 1937 steel castings constituted only about 3.4% of the combined production of finished rolled steel products and steel castings in the United States, yet over 300 foundries were in operation producing highly diversified products essential to the operation of many other industries.

Dendritic Structure of Cast Steel. — Steel cast in any form is essentially dendritic in structure.   Upon solidification it undergoes the same processes of dendrite formation and chemical segregation as described in Chapter II for solid solution type alloys.   The dendrites may be greatly elongated as in the surface layers of most ingots and many castings, or they may be more or less equiaxed as usually found in the central portions of cast structures.

Carbon, manganese, silicon, and phosphorus are the principal elements in solid solution in the iron at high temperatures just after solidification.   Phosphorus is generally conceded to be one of the principal elements causing the relatively permanent type of segregation which is responsible for dendritic structures.   However, silicon,

nickel, and other alloying elements are also subject to segregation. Chemical analysis generally shows a difference in the compositions of various portions of ingots and large castings, the carbon, sulphur, phosphorus, and alloying elements being somewhat higher in the

FIG. 165. — MACROSTRUCTURE OF A STEEL CASTING. (Etched in a boiling solution of 50% HCl-50% H₂O, then repolished lightly.) X3. (*Courtesy of H. S. Jerabek.*)

central and upper portions which solidify last. This gross chemical segregation should not be confused with the microsegregation existing in the minute structural parts of the individual dendrites. It is believed that this latter form of segregation has a marked influence on the mechanical properties, and one of the objectives of heat treatment of steel castings is the homogenization of the cast microstructure.

The dendritic structure of a cast steel is illustrated in Fig. 165,

which was prepared by deep etching followed by very light polishing to improve the contrast.   In this casting the columnar surface dendrites extend through the entire section, which is slightly less than one inch thick, forming cleavage planes at the corners.   (Only one-half of the cross section is shown.)   In many cases the dendrites are equiaxed in the interior and it is often difficult to produce evidence of dendrite formation in the macrostructure of steel castings.   (See Fig. 172 for example.)

It is agreed that homogenization of cast structures by mechanical work and heat treatment improves the mechanical properties.   Only the latter method is available, of course, for steel castings and most welded structures.   It is interesting to note that in many cases visible evidence of the original dendritic structure, as revealed by etching, persists after considerable mechanical work has been done on the steel.   Bars of high quality tool and alloy steels sometimes retain dendritic patterns in their final heat treated condition.   It has not been proved that such steels are inferior to those having no trace of dendritic structure.

The subject of dendrites in steels has been discussed at length by Sauveur [3], Keshian [4], and Sauveur and Reed.[5]

Microstructure of Cast Steel. — During the initial cooling of the solid casting certain changes take place which play an important part in determining the microstructure.   The large dendritic units which prevail just after solidification are austenitic in structure. They tend to break down into smaller austenitic units of variable orientation while cooling down to the Ar_3 temperature.   This is known as the granulation period.   The resulting austenite grains, while smaller than the original dendrites, are usually coarse compared with those of corresponding wrought steels.   Upon further cooling through the critical temperature range the austenite grains undergo transformation as already described in Chapter VII; thus the final microstructure, while more or less comparable to that of a normalized wrought steel, is normally coarser both from the standpoint of the size of the microconstituents (ferrite and pearlite) and the size of the austenitic units from which they form.

The microstructure of a low carbon steel casting containing 0.18% carbon is shown in Fig. 166.   The wide ferrite bands mark

the boundaries of former austenite grains. Additional ferrite has separated profusely within the grains as the casting cooled between the Ar_3 and Ar_1 temperatures. The dark areas are pearlite which transformed from the remaining austenite at Ar_1. (See also Figs. 167 to 174.)

Solidification of Steel Castings. — In addition to these metallographic considerations which give us an understanding of the structural condition of cast steel, several practical foundry problems are involved in the solidification of steel castings. These include the selection of suitable methods of gating and feeding, control of metal pouring temperature and speed of pouring, the type of mold and its condition, and the use of chills. Some of the difficulties encountered are formation of visible and microscopic shrinkage cavities, cracking or hot tears due to cooling stresses, failure to fill the mold because of lack of fluidity, porosity caused by mechanical inclusion of gases, inclusion of slag and of sand washed from the mold, surface defects, etc.[6]

FIG. 166. — STEEL CASTING CONTAINING 0.18% CARBON. (Etched with picral.) X50.

While some of these difficulties occur in ingot making, the production of intricate castings requires extra care to forestall failure. The greatly improved reliability and the high mechanical properties now available in steel castings are an indication of the progress being made. A series of papers by Briggs and Gezelius[7] contains probably the most complete quantitative data now available on the solidification and contraction of steel castings. It is their opinion, as well as that of many other investigators, that directional solidification should replace uniform cooling of the entire casting. This simply means that portions which are remote from the sources of molten metal should be made to solidify

first, followed by progressive freezing from these sections towards the risers which should contain sufficient molten metal to feed the last portions undergoing liquid shrinkage. Although uniformity of section is always preferred over abrupt changes, it is easy to see that uniformity in itself will not give the desired result. In fact, the tapered ingot molds used in modern ingot practice are an excellent example of non-uniform sections used to attain directional solidification. In foundry practice some of the devices used for this purpose are: the placing of chills to hasten solidification in certain sections, selection of proper locations for gates and risers, and in some cases the manipulation of the mold after pouring by tilting or inverting.[8, 9]

Briggs and Gezelius[7] have obtained quantitative data on contraction in the mold, another important phase of the solidification of steel castings. The temperature interval between final solidification and room temperature is greater for cast steel than for cast iron. The contraction occurring in this interval is likewise greater for steel castings, therefore special provision must be made for freedom in contraction in order to avoid high cooling stresses. Although permanent metal molds are sometimes used for molding cast iron, such restriction in the case of steel castings would probably result in cooling stresses of sufficient magnitude to rupture all but the simplest shapes. In fact, the molds and cores used for long or intricate steel castings must be more or less collapsible to prevent hot tears. It is obvious that bosses and other protruding parts hinder free contraction. When this occurs it is, of course, advisable to avoid steel compositions known to be " tender " or " hot-short " at red heats.

Another interesting observation by the same investigators is that the cooling rate of castings in green sand molds is only slightly higher than that in dry sand molds. Both green sand[10] and dry sand molds are used successfully. An outstanding example of the latter practice is the production of automobile crankshafts in molds built up from sixteen individual dry sand and " core " sections.[11] The sand used in either case must be highly refractory to withstand the high pouring temperature of steel castings. A special patented process makes use of portland cement as a binder. Cemented molds dry in air at room temperature until they are as hard as bricks, therefore flasks are not needed to retain the metal upon pouring.[12]

Design of Steel Castings. — The most successful applications of steel castings are naturally those in which the design is adapted to the casting characteristics and physical properties of cast steel. A member which has been properly designed for cast iron cannot ordinarily be cast in steel if full advantage is to be taken of the properties of steel. Likewise members which have been forged or otherwise fabricated from wrought steel require considerable alteration if cast steel is to be applied. In the cases of many railroad and rolling mill castings which have been made in large quantities for several years the designs are, of course, well adapted to foundry conditions, but in the present competition between foundry products and fabricated steel the design must be specialized for each type of construction.

A few of the general considerations in the design of steel castings are: equalization of sections and gradual transitions between unlike sections; provision for flow of metal to such thin sections as are required; elimination of sharp changes in direction of stressed members to avoid stress concentration; elimination of projections on long castings because of their tendency to restrict contraction and cause cracking; simplification of core requirements; and the use of relief cores to reduce the mass of metal at the junction of several sections. Lorenz [13] and Wheeler [14] have described many design features of steel castings.

Welding of Steel Castings. — Before the development of welding methods capable of giving uniformly dependable results, the repair of defects in steel castings was a very questionable procedure. It is now permissible to weld many types of castings. When the carbon content is under 0.25% no great difficulties are involved [15]; on the other hand, high carbon and alloy steels tend to air harden and crack at the weld. It is possible to weld the higher carbon castings successfully when proper precautions are taken in preparing the section, preheating the casting, applying the weld metal, peening the deposited metal, and annealing the entire casting without permitting the weld to cool down before heat treatment. [16]

Manufacturers of steel castings have long considered welding as a competitor. It is unquestionably true that parts and assemblies which were once cast either in iron or steel are now being fabricated by means of welding of various wrought products such as sheet,

plate, shapes, and forgings. Both steel and iron foundries are meeting this situation by improving their product. The inherent advantage of castings for complicated sections cannot be discounted, and it is believed that many designs of the future will include both cast and wrought parts joined by welding. Progress is being made in this direction using steel castings with less than 0.40% carbon welded to other castings and to rolled products.[17, 18] The ductility of the assembly can be improved by annealing.

Carbon Steel Castings. — Although an important part of the total production of steel castings consists of alloy steels, carbon steels are the mainstay of the foundry industry. The compositions used are similar to those of wrought carbon steels except that the silicon and manganese are higher. These elements are necessary to " kill " the steel and prevent gassing from the metal in the mold. This is more essential in castings than in ingots because there can be no welding shut of blowholes as often occurs during rolling or forging.

A comparatively few grades of carbon steels constitute the bulk of the production and of these the so-called medium carbon or regular grade is the most important. There has been considerable difference in opinion as to the composition range of " medium carbon " castings. Recently The American Foundrymen's Association has recommended the wide range from 0.12–0.49% carbon, thus low carbon castings contain less than 0.12% and high carbon castings over 0.49% carbon according to this definition. The other chemical qualifications for plain carbon steel castings are 0.50–1.00% manganese, 0.20–0.75% silicon, 0.05% maximum phosphorus, and 0.06% maximum sulphur.[19]

Adherence to this classification would undoubtedly reduce confusion when discussing these castings; however, there is need for subdivision of the medium carbon range when specific properties and applications are considered. In " Metals Handbook "[20] the three carbon ranges 0.15–0.30%, 0.30–0.40%, and 0.40–0.60% carbon are given as a basis for heat treating practice. This arrangement will be used in Table LI to follow.

The high carbon range has also been subdivided for convenience in classification by foundries specializing in such castings; for ex-

ample, one manufacturer of cast steel rolls considers " low carbon " to be 0.50–0.80%, " medium carbon " 0.80–1.40%, and " high carbon " 1.40–2.60%.

Heat Treatment of Steel Castings. — Normalizing, annealing, quenching, and drawing are the principal treatments applied to steel castings. Sims [21] has studied the effect of heat treatment on medium manganese cast steels containing 0.30–0.40% carbon and 1.25–1.50% manganese. His photomicrographs, reproduced here as Figs. 167 to 174, illustrate the principles of grain refinement of carbon as well as low alloy cast steels.

The cast or raw structure of a 1 inch section is shown in Fig. 167. The cooling rate in a sand mold approximates the rate obtained upon air cooling or normalizing, consequently the pearlite in this figure and those following contains less than the equilibrium carbon content for the eutectoid, and the relative volume of the dark areas is greater than would be found in fully annealed wrought steel of the same carbon content. The volume of the ferrite is correspondingly less. The large primary grains of Fig. 167 are outlined by the coarse ferrite network. However, each of the large pearlite areas is composed of several smaller units which formed when the austenite grains transformed to pearlite at Ar_1.

Heating above the critical temperature range and air cooling produced grain refinement as shown in Fig. 168. Although recrystallization has been quite complete, a duplex condition of grain sizes has been established. This uneven response to heat treatment is attributed to dendritic segregation. It has been found that higher heat treating temperatures tend to homogenize the cast structure and give a coarse but more uniform grain size as in Fig. 169. The comparatively coarse structure produced by this preliminary high temperature normalizing treatment may then be refined by a second treatment just over the critical temperature range, giving the structure shown in Fig. 170. This procedure is known as double normalizing.

The effect of vanadium on grain refinement, and consequently upon the toughness of steel castings, is illustrated by Fig. 171, whose composition and heat treatment are the same as for Fig. 170 except for the addition of 0.1% vanadium.

FIG. 167
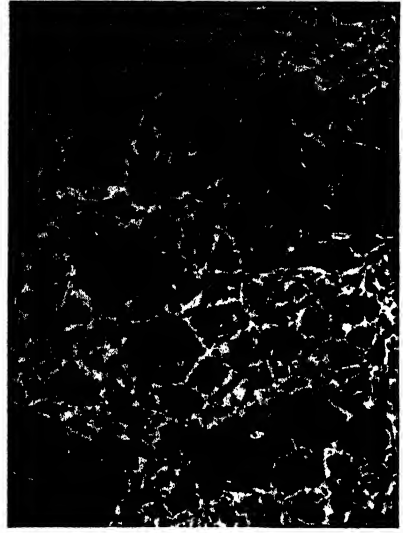
FIG. 168
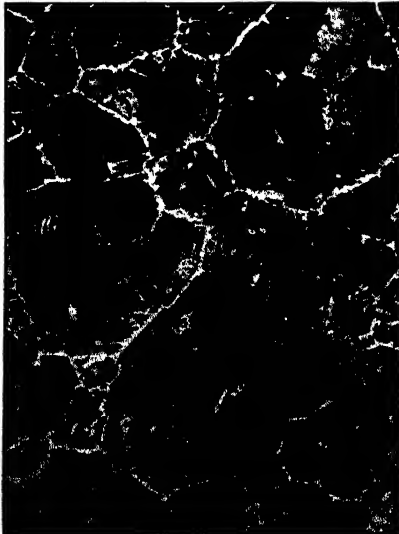
FIG. 169
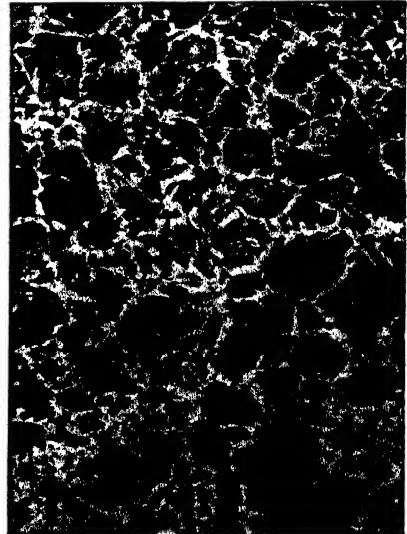
FIG. 170

FIGS. 167 TO 174 REPRESENT THE STRUCTURES OF MEDIUM MANGANESE STEEL CAST-INGS AFTER VARIOUS TREATMENTS. ALL CASTINGS CONTAINED 0.30–0.40% CARBON AND 1.25–1.50% MANGANESE. (*Courtesy of C. E. Sims, Metal Progress.*[21])

FIG. 167. — UNANNEALED OR RAW STRUCTURE AS CAST IN A 1 INCH SECTION. X100.

FIG. 168. — AIR COOLED AFTER 1 HOUR AT 1550° F. X100.

FIG. 169. — AIR COOLED AFTER 1 HOUR AT 1650° F. X100.

FIG. 170. — STRUCTURE OF A 1 INCH SECTION AFTER A DOUBLE NORMALIZING TREATMENT. X100.
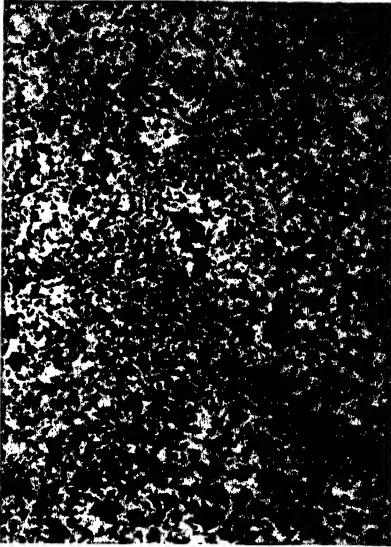
FIG. 171
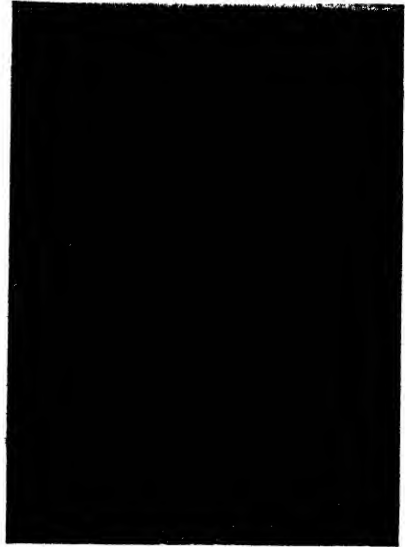
FIG. 172
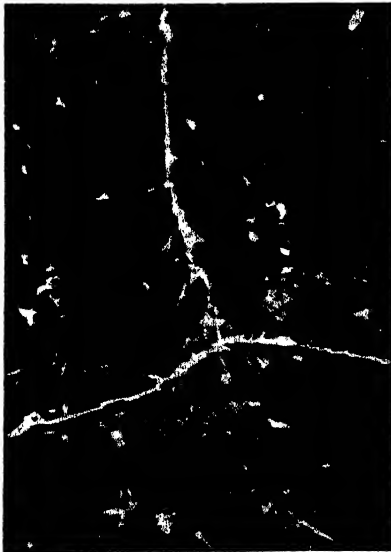
FIG. 173
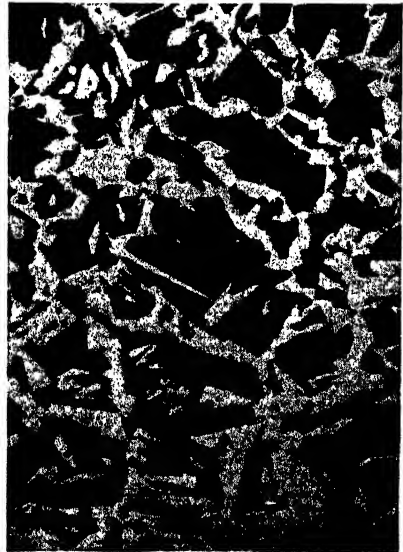
FIG. 174

FIG. 171. — SAME AS FIG. 170 EXCEPT FOR THE ADDITION OF 0.1% VANADIUM. X100.

FIG. 172. — MACROSTRUCTURE OF A RAW CASTING 6 INCHES IN SECTION. X1.25.

FIG. 173. — MICROSTRUCTURE AT THE CENTER OF A RAW CASTING 6 INCHES IN SECTION. X25.

FIG. 174. — SAME AS FIG. 173 AIR COOLED AFTER 6 HOURS AT 1650° F. X100.

The structure of heavy sectioned castings may be very coarse, as shown in Figs. 172 and 173 at X1.25 and X25 respectively. The low cooling rate during solidification of this 6 inch section increased the tendency to segregate, consequently heating for 1 hour at 1650° F. followed by air cooling was insufficient to homogenize the casting, resulting in a duplex grain structure. After 6 hours at 1650° F. the uniform grain structure of Fig. 174 was obtained. This structure may be further refined by a second treatment just above the critical temperature if desired. The comparatively large size of the ferrite units in Fig. 174 compared with Fig. 170 is the result of the relatively slow transformation of austenite to ferrite at the center of the 6 inch section.

Almost all castings are inferior to the corresponding wrought metals in their impact toughness. One of the objectives of heat treatment of steel castings is improvement in toughness. The coarse grain formation of raw steel castings is responsible for coarse fractures and poor impact values. Annealing generally improves this condition but the ferrite may still be present as a thick network in medium carbon steels, especially if cooling is very slow. Normalizing or quenching, followed by suitable reheating, tends to give finer structures with higher impact values. Grotts [22] has presented data on side frames for freight car trucks of the following average composition: 0.29% C, 0.70% Mn, 0.38% Si, 0.03% P, 0.03% S.

	Full Anneal	Full Anneal and Normalize	Full Anneal, Quench and Draw
Tensile strength, lb./sq. in. ...	65,000	67,000	70,000
Yield point, lb./sq. in.	38,000	40,000	46,000
Elongation in 2 in., %	29	26	25
Reduction of area, %	40	45	52
Izod, ft. lb.	12	22	40
Microstructure	Coarse separation of ferrite and pearlite	Fine but distinct ferrite and pearlite	Sorbitic structure

Similar results were obtained by Knox [23] by means of quenching and drawing treatments on 1½ inch square bars from 1880 heats of 0.30% carbon basic open hearth and acid electric furnace steels.

	As Cast	Annealed	Heat Treated High Draw	Low Draw
Tensile strength, lb./sq. in.	74,100	75,000	80,000	125,000
Yield point, lb./sq. in.	37,100	41,500	50,000	97,000
Elongation in 2 in., %	19.5	24.5	30	10
Reduction of area, %	31.0	46.2	65	20
Brinell hardness	160	145	155	250
Endurance ratio	0.40	0.44	0.50	0.50
Izod, ft. lb.	16	20	48	15

In a large number of cases it is not possible to apply drastic quenching treatments to castings because of their intricate shapes or their large masses; consequently annealing or normalizing treatments such as those described earlier in this section are widely used. Armstrong [24] has carried out extensive annealing and normalizing heat treatments on 1 carbon and 25 alloy steels covering a wide range of compositions with carbon contents from 0.10 to 0.38%. The average mechanical properties of the 26 heats after nine different heat treatments are summarized in Table XLIX.

The double annealing treatments D, E, and F gave an increase in ductility, yield point, and impact toughness compared with the single

TABLE XLIX. AVERAGE PROPERTIES OF TWENTY-SIX HEATS
OF ALLOY STEEL CASTINGS

Treatment No.	El. in 2 in.,%	Red. of Area, %	Yield Point lb./sq. in.	Tensile Strength lb./sq. in.	Izod ft. lb.	Treatment
A	10.1	15.5	74 190	100 650	6.6	As cast
B	24.7	46.0	61 730	94 290	27.3	Annealed 1650° F.
C	25.5	46.4	60 780	93 400	—	Annealed 1700° F.
D	26.3	48.7	63 860	93 730	—	Double annealed 1650°–1650° F.
E	26.1	48.3	64 350	94 430	38.6	Double annealed 1650°–1525° F.
F	27.0	49.7	61 560	91 770	—	Double annealed 1700°–1500° F.
G	26.4	49.1	67 120	93 660	—	Normalized 1650° F. — Annealed 1525° F.
H	24.9	51.5	67 590	95 590	31.5	Normalized 1650° F. — Tempered 1200° F.
J	25.9	52.8	69 500	93 480	44.6	Double normalized 1650°–1650° F. — Tempered 1200° F.
K	26.2	52.4	60 090	89 920	36.0	Double normalized 1700°–1525° F. — Tempered 1300° F. — 6 hrs.

annealing treatments B and C. Single normalizing followed by tempering, treatment H, is one of the most frequently used commercial treatments. Armstrong's results indicate good ductility, toughness, and high strength after this treatment. Double normalizing and tempering at 1200° F., treatment J, gave the highest combination of mechanical properties. Treatment K, involving tempering for 6 hours at 1300° F., is essentially a spheroidizing treatment to be used to increase the ductility of hard alloy steels. A longer time at 1300° F. is generally necessary, however, to produce full spheroidization.
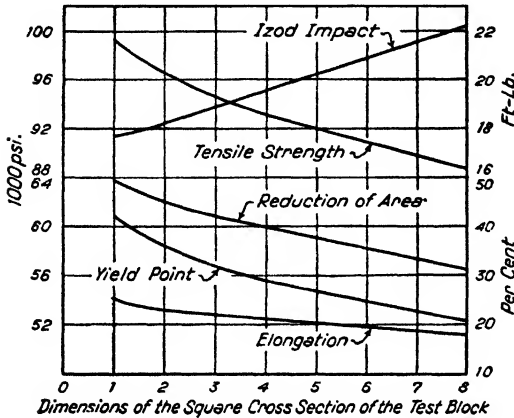
FIG. 175. — EFFECT OF MASS UPON THE MECHANICAL PROPERTIES OF MEDIUM MANGANESE STEEL AT THE CENTER OF THE TEST BLOCK. (*Courtesy of Charles W. Briggs and Roy A. Gezelius, Trans. Am. Soc. for Metals.*[26])

It is clear that some treatment involving heating through the critical temperature range is highly desirable, and that one of several procedures may be chosen depending upon the nature and requirements of the casting.

It should not be inferred that all steel castings are heat treated. A large tonnage of light, uniform sectioned, low carbon steel castings has been used successfully in the as-cast state by the railways. Such castings compete with malleable cast iron for use in small parts.[25]

Effect of Mass on the Properties of Steel Castings. — The properties of steel castings depend not only upon composition and heat treatment but upon section size and a number of other variables which may be summarized under the heading of foundry practice.

Briggs and Gezelius [26] have studied the effect of mass on the properties of steels cast in sections varying from ½ inch to 8 inches square.

The tensile properties of test specimens taken from the center

sections of castings decreased as the size of the section increased. The results for a medium manganese steel, as given in Fig. 175, indicate that only impact toughness increased in the larger sections. It should be kept in mind that the center properties are lower than those at the surface in large sections.

Briggs and Gezelius consider the changes in mechanical properties to be associated with the coarsening of the microstructure caused by slower cooling, and the lowering of the carbon content and density as the section size increases. (The carbon contents in the centers of 1 inch and 8 inch medium manganese steel castings were 0.34 and 0.30% respectively. A similar drop was observed in a 0.25% carbon steel. The corresponding densities were 7.838 and 7.811 for the carbon steel castings.)

As in wrought steels the effect of mass is of considerable importance in heat treatments requiring rapid cooling. Plain carbon cast steels are difficult to harden in large sections, even in the higher carbon range. Many alloy compositions are available having much greater hardenability, as illustrated in Table L by the data for a chromium-molybdenum steel as reported by Heaslett.[27]

TABLE L.   BRINELL HARDNESS OF WATER QUENCHED CAST STEELS

Size of section	Plain carbon cast steel with 0.50% C		Alloy cast steel containing 0.40% C, 0.80% Cr, 0.25% Mo, 0.75% Mn, 0.30% Si	
	surface	center	surface	center
1 inch	592	321	601	578
2 inch	401	285	555	514
3 inch	341	248	534	477
4 inch	293	229	495	429
5 inch	255	223	461	401

Mechanical Properties of Steel Castings. — Published data for representative alloys have been assembled in Table LI in order to give a cross section of both the compositions and properties of commercial castings. Except for certain special cases the alloy grades were selected with carbon contents within the range 0.30–0.40%. In most instances the results given are test data from one or more heats of the average composition given. It is not likely

TABLE LI. COMPOSITIONS AND PROPERTIES OF STEEL CASTINGS

No.	Type	C	Mn	Si	Special Elements	Heat Treatment*	Tensile Strength lb./sq. in.	Yield Point lb./sq. in.	Elongation in 2 in. %	Reduction of Area %	Impact** ft. lb.	References
1	Low Carbon	0.13	0.96	0.36	—	As cast	62 000	33 000	13.1	14.2	I20	28
2	"	0.13	0.96	0.36	—	1690° Ann.	68 000	38 000	25.0	38.3	—	28
3	"	0.17	0.67	0.23	—	1650° Ann.	64 000	35 000	28.5	40.2	—	28
4	"	0.20	0.73	0.40	—	1650° Ann.	69 000	43 000	27.2	39.7	—	28
5	"	0.21	0.94	0.19	—	1560° N. 1025° D	80 000	56 000	28.0	45.0	—	28
6	"	0.24	0.71	0.33	—	Ann.	73 100	41 900	30.5	48.5	—	28
7	"	0.25	0.68	0.32	—	1650° N, 1525° N, 1000° D	76 000	43 000	31.7	56.0	C34	28
8	Med. Carbon	0.30	—	—	—	Ann.	75 000	41 500	24.5	46.1	I20	29
9	"	0.30	0.62	0.18	—	1740° N, 930° D	83 000	50 000	16.2	23.9	C12	28
10	"	0.30	0.70	0.28	—	1650° N, 1200° D	79 000	45 000	31.0	47.0	I34	24
11	"	0.37	0.79	0.40	—	1650° N, 1290° D	88 000	49 000	21.4	28.3	—	28
12	High Carbon	0.42	0.65	0.28	—	1740° N, 930° D	88 000	52 000	18.2	20.4	C5.5	28
13	"	0.48	0.68	0.41	—	1650° N, 1290° D	91 000	54 000	21.3	29.5	—	28
14	"	0.69	1.03	0.25	—	1510° Ann.	110 000	47 000	8.1	7.8	—	28
15	"	0.86	0.90	0.27	—	1470° Ann.	108 000	51 000	4.0	3.6	—	28
16	Mn	0.32	1.16	0.35	—	1650° Ann., 1550° N, 700° D	82 000	46 000	30.0	47.5	—	30
17	"	0.35	1.38	0.33	—	1650° N, 1290° D	84 000	47 000	24.5	35.1	—	28
18	"	0.32	1.42	0.40	—	1650° N, 1250° D	83 000	55 000	31.5	55.7	C16.5	28
19	"	0.37	1.44	0.28	—	1650° N, 1250° D	94 000	58 000	24.5	49.7	C4.5	28
20	"	0.33	1.50	0.38	—	1650° N, 1250° D	89 000	51 000	28.0	58.3	C13.5	28
21	"	0.35	1.51	0.31	—	1650° Ann., 1550° N, 700° D	97 500	55 800	25.0	53.0	I23	30
22	"	0.36	1.67	0.29	—	1650° Ann., 1550° N, 700° D	107 000	57 500	21.5	45.0	I20	30
23	"	0.36	1.75	0.29	—	1650° Ann., 1550° N, 700° D	109 000	57 000	22.0	46.0	I23	30
24	"	0.36	1.83	0.29	—	1650° Ann., 1550° N, 700° D	115 000	65 000	20.0	43.0	—	30
25	V	0.38	0.91	0.32	0.18 V	Ann., N, and D	91 500	58 500	25.5	50.5	I32	30
26	"	0.30	0.75	0.35	0.20 V	1700° N	98 000	62 000	24.0	39.7	—	28
27	"	0.40	0.75	0.33	0.23 V	Double N and D	87 700	59 200	27.2	55.4	—	27
28	Mn-V	0.32	1.38	0.41	0.10 V	1650° Ann., 1550° N, 700° D	98 000	70 200	28.0	57.6	I59	30
29	" "	0.35	1.40	0.40	0.10 V	Double N	104 000	79 000	27.0	54.7	I49	28
30	" "	0.35	1.40	0.40	0.10 V	N and D	99 000	71 900	27.7	56.4	I52	31
31	Mn-Ti	0.30	1.55	0.40	Ti treated	1600° N, 1050° D	99 000	73 500	31.0	60.0	I49	32
32	Mo	0.35	0.71	0.29	0.34 Mo	Ann., N, and D	83 200	58 100	21.0	33.0	I28	30
33	"	0.38	0.93	0.24	0.36 Mo	1650° N, 1200° D	93 750	65 500	23.5	48.5	I28	24
34	"	0.26	—	—	0.61 Mo	1600° N, 1300° D	86 700	61 000	27.0	60.0	—	33

No.	Type	C	Mn	Si	Alloy	Heat Treatment						
35	Mn-Mo	0.33	1.60	—	0.35 Mo	1650° N, 1250° D	96 400	68 000	25.5	57.5	—	27
36	"	0.35	1.50	0.40	0.15 Mo	Double N	114 050	69 800	21.0	41.6	I119	34
37	"	0.35	1.35	—	0.35 Mo	N and 1250° D	96 000	68 000	25.5	51.7	—	29
38	"	0.30	1.20	—	0.20 Mo	1600° N, 1300° D	84 000	60 000	27.0	55.0	—	33
39	Mn-Mo-V	0.36	1.17	0.48	0.33 Mo, 0.10 V	Double N	133 450	90 750	12.0	25.4	125.5	34
40	" "	0.35	1.50	0.40	0.15 Mo, 0.10 V	Double N	118 650	76 050	23.0	39.1	130.5	34
41	Cu	0.28	0.75	0.21	0.82 Cu	1650° N, 1200° D	81 750	59 500	28.0	55.0	140	24
42	"	0.33	0.74	—	1.02 Cu	1550° N, 1000° D	107 000	79 000	20.0	39.7	—	28
43	"	0.31	0.75	0.42	1.21 Cu	1650° N.	97 500	68 000	25.5	51.0	C17	35
44	"	0.31	0.75	·0.42	1.21 Cu	1650° N, 930° Ppt.-3 hrs.	110 500	82 500	21.0	45.8	C12	35
45	Cu-Mn-Si	0.15	1.43	1.10	0.34 Cu	1650° N, 1200° D	84 500	58 000	30.5	63.0	124	24
46	" " "	0.15	1.08	1.09	1.89 Cu	1650° Ann.	89 750	69 550	29.7	57.2	—	24
47	" " "	0.20	1.16	0.70	1.75 Cu	1650° Ann.	93 500	80 000	23.5	41.5	—	24
48	" " "	0.11	1.04	1.25	1.74 Cu	1740° Ann.	81 500	64 000	31.0	56.0	—	36
49	" " "	0.11	1.04	1.25	1.74 Cu	1740° Ann.,1580° N, 915° Ppt.	106 500	88 500	25.0	53.0	—	36
50	Cr	0.35	0.67	0.37	0.77 Cr	1650° N, 1250° D	97 000	56 000	17.5	26.0	—	28
51	"	0.36	0.74	0.41	0.83 Cr	Ann., N, and D	95 000	55 000	18.0	31.0	I29	30
52	"	0.30	0.80	0.40	1.00 Cr	Double N	94 500	60 400	27.5	54.7	I38	34
53	Cr-V	0.30	0.80	0.40	1.00 Cr, 0.10 V	Double N	94 300	64 850	27.5	57.1	I59	34
54	Cr-Mo	0.39	0.81	—	0.69 Cr, 0.43 Mo	N and 1250° D	103 000	73 000	19.0	40.0	—	29
55	" "	0.30	—	—	0.80 Cr, 0.20 Mo	1550° N. 1250° D	110 000	80 000	20.0	40.0	—	33
56	Mn-Cr	0.38	1.38	0.47	0.63 Cr	1650° Ann...1550° N, 700° D	121 900	72 450	22.0	53.0	—	30
57	Mn-Cr-V	0.34	1.51	0.29	0.47 Cr, 0.16 V	1650° N, 1200° D	112 250	86 750	16.0	41.5	I7	24
58	Mn-Cr-Mo	0.34	1.19	—	0.96 Cr, 0.34 Mo	1600° N, 1300° D	105 800	77 450	23.0	55.0	—	33
59	Ni	0.18	0.85	—	2.19 Ni	Double N and D	80 900	51 300	30.0	60.1	I59	37
60	"	0.35	0.88	0.31	1.54 Ni	1510° N to 1110°, Fce. Cool	103 000	61 000	16.0	32.5	—	28
61	"	0.38	1.07	0.19	2.20 Ni	1650° N, 1200° D	103 100	71 000	22.0	46.0	I30	24
62	"	0.35	0.80	—	3.05 Ni	N and D	98 000	67 700	27.0	49.0	I49	38
63	Mn-Ni	0.34	1.48	0.44	0.47 Ni	1650° Ann...1550° N, 700° D	94 000	54 500	28.5	58.0	—	30
64	" "	0.32	1.10	0.31	1.10 Ni	N and 1250° D	90 400	63 000	25.0	40.0	—	29
65	" "	0.31	1.17	0.37	1.22 Ni	1600° N, 1200° D	94 550	76 400	27.5	50.0	—	24
66	" "	0.27	1.41	—	1.40 Ni	N and D	99 200	61 000	24.5	57.5	I41	37
67	" "	0.31	1.60	—	1.41 Ni	N and D	101 700	68 400	26.5	59.5	I65	37

* N = Normalize. D = Draw.   ** I = Izod. C = Charpy.

TABLE LI. COMPOSITIONS AND PROPERTIES OF STEEL CASTINGS (*Continued*)

No.	Type	C	Mn	Si	Special Elements	Heat Treatment*	Tensile Strength lb./sq. in.	Yield Point lb./sq. in.	Elongation in 2 in.,%	Reduction of Area,%	Impact** ft. lb.	References
68	Ni-V	0.19	0.64	—	1.62 Ni, 0.10 V	Double N and D	83 500	57 000	27.0	52.0	—	37
69	"	0.25	0.91	0.33	1.55 Ni, 0.11 V	1650° Ann., 1550° N, 700° D	91 000	61 000	28.0	57.0	I57.5	30
70	"	0.28	1.00	—	1.50 Ni, 0.10 V	N and D	96 120	69 300	27.8	56.5	I54.5	31
71	"	0.29	0.89	—	1.59 Ni, 0.11 V	Double N and D	94 400	65 300	27.3	55.7	I56	37
72	"	0.30	0.81	—	1.52 Ni, 0.11 V	N and D	96 000	66 200	27.5	55.0	I56	38
73	Ni-Mo	0.33	0.70	—	1.37 Ni, 0.33 Mo	N, 1250° D	91 000	60 500	24.8	54.4	—	29
74	"	0.36	0.98	0.17	1.92 Ni, 0.26 Mo	1650° N, 1200° D	102 750	77 000	21.0	46.5	I30	24
75	"	0.30	—	—	1.50 Ni, 0.30 Mo	1600° N, 1200° D	90 000	60 000	22.0	45.0	—	33
76	Ni-Cr	0.35	0.80	0.40	1.30 Ni, 0.94 Cr	N, 1250° D	102 000	67 500	21.0	41.0	—	29
77	"	0.37	0.79	0.41	1.38 Ni, 0.89 Cr	Ann., N, and D	103 000	66 000	20.0	39.0	I37	30
78	"	0.37	0.76	0.39	1.33 Ni, 0.98 Cr	1650° N, 1250° D	104 000	66 000	22.5	42.0	—	28
79	"	0.35	0.82	0.36	1.95 Ni, 0.66 Cr	1500° N, 1200° D	106 000	93 000	23.5	52.0	I37	28
80	"	0.30	—	—	2.00 Ni, 0.90 Cr	1550° N	110 000	75 000	20.0	40.0	—	33
81	"	0.51	0.71	0.39	2.85 Ni, 0.87 Cr	1550° N, 1250° D	108 000	72 000	17.5	24.8	C8	28
82	Ni-Cr-Mo	0.30	—	—	2.00 Ni, 0.85 Cr, 0.25 Mo	1650° N, 1200° D	114 000	89 000	21.0	50.0	—	33
83	"	0.35	0.85	—	1.75 Ni, 0.75 Cr, 0.35 Mo	N, 1250° D	118 000	90 000	18.0	35.1	—	29
84	Ni-Cr-Mn-Mo	0.35	1.25	—	1.50 Ni, 0.90 Cr, 0.30 Mo	1650° N, 1100° D	138 000	118 900	16.0	45.0	—	33
85	"	0.34	1.58	—	1.22 Ni, 0.71 Cr, 0.32 Mo	N, 1250° D	125 000	92 000	22.7	51.2	—	29

* N = Normalize. D = Draw.   ** I = Izod.   C = Charpy.

that best practice was used in every case, and it is entirely possible that some of the values reported do not reflect the full possibilities of a given composition. Although detailed information is not always given in the original source, it is probably true that in the majority of cases the test data were obtained from test coupons cast separately or attached to the castings, while in some instances the test samples were machined from the casting. The properties are usually representative of relatively light sections. (Minimum specification requirements for many grades of steel castings will be found in the American Society for Testing Materials Standards.)

It will be noted that in most cases the final heat treatment consisted of normalizing followed by reheating or drawing to approximately 1000°–1250° F. This treatment was selected because it is widely used in practice and more data were available than for any other single treatment. It is true that this method of comparison works to the disadvantage of certain alloy compositions which require quenching to develop their best properties. (In certain cases data on mechanical properties in the annealed and in the quenched and drawn state are available in the original references.)

Low Carbon Steel Castings. — The casting characteristics and mechanical properties of the low carbon grades, Nos. 1–7 in Table LI, are inferior to those of the medium carbon grades. Their application depends, in many instances, upon their weldability, elevated temperature properties, magnetic characteristics, and suitability for carburizing. Typical applications are annealing boxes, furnace parts, ladles, motor and other electrical castings, large carburized gears, and miscellaneous frames, brackets, and housings for machinery and railway uses where the superior strength properties of the medium carbon and alloy grades are not required. In many large castings the requirements for rigidity and ductility and ease of repair by welding are much more important than high strength and wear resistance.[39] The low carbon grades often compete with malleable and gray cast iron for small parts.

Medium Carbon Steel Castings. — By varying the carbon content in the range 0.30–0.40% and by proper heat treatment, castings may be produced having properties suitable for a wide variety of products. If the wider limits of 0.24–0.42% carbon are taken to

define the medium carbon range, the greater part of the production of steel foundries will be included.

The application of X-ray methods of inspection has been of great assistance in the development of sound steel castings capable of competing favorably with other iron and steel products including gray, malleable, pearlitic malleable, and alloy gray irons, as well as wrought steels.[16, 40, 41]

The ordinary mechanical properties of medium carbon cast steels (Nos. 8–11 in Table LI) are similar to those of wrought steels of the same carbon contents except that they have lower ductility, particularly as measured by per cent reduction of area.

The proper use of liquid quenching treatments followed by suitable drawing improves the properties of cast carbon steels to such an extent that they may sometimes be used where alloy cast steels would otherwise be required. In general, the drawing temperatures used for steel castings are higher than for forgings in order to provide satisfactory ductility and toughness. Gregg's data[33], representing the results obtainable with a 0.30% carbon steel when the best foundry and heat treating practice is applied, include the following properties measured after quenching and drawing at 1200° F.: tensile strength 86,000 lb. per sq. in., yield point 60,000 lb. per sq. in., elongation in 2 inches 25%, reduction in area 60%, Brinell hardness 250, and Izod impact 43 ft. lb.

High Carbon Steel Castings. — The high carbon cast steels, 0.40% carbon and over, are somewhat higher in strength but lack the toughness and ductility of the medium carbon grades. (See Nos. 12–15, Table LI.) However, their greater hardness makes them more suitable for certain gears, wheels, rolls, dies, tractor shoes, and many other parts subject to abrasive wear. Rolls for steel mills and other purposes are often very high carbon steel castings, with or without alloying elements. (Chill cast iron is also used, especially for plain rolls.)

Herington[42] describes a 0.40% carbon steel casting weighing 34,-000 pounds which is used as the casing for a large dredge pump. These castings are annealed at 1600° F. before machining. The service requirements include resistance to abrasion, impact, and corrosion. Carbon steel castings of much larger size, up to 165 tons

and limited only by transportation facilities, have been produced for use in rolling mills, special machine tools, and large engines.[43]

Alloy Steel Castings. — About 15% of the present production of steel castings are alloy grades, a large percentage of which are melted in electric arc furnaces. In order to produce a high grade product in which the full benefit of the alloy addition will be realized, extra care must be taken in all stages of production. As an example, the tendency of many compositions to air harden, especially when cast in light sections, makes it advisable to remove the castings from the mold while still at a red heat and charge them into an annealing furnace. After slow cooling in the furnace the castings may be reheated to 1000°–1200° F. and the gates and risers burned off with a torch.

The many compositions available and the diversity of mechanical properties which can be produced by heat treatment cover as wide a range as in wrought alloy steel practice. In addition to the large and important group of medium alloy steels (similar to the wrought S.A.E. steels), both low alloy and high alloy steel castings are made. In the low alloy field a successful precipitation hardening composition has been introduced. In the high alloy field most of the stainless and heat resisting compositions can be made in cast form.

Mn. — Corresponding to the " semi-alloy " wrought steels such as the S.A.E. T1300 series, the medium manganese cast steels fill a need for castings with higher tensile properties and toughness than obtainable with straight carbon steels, and at a cost only moderately greater than that of carbon steels. The usual range of compositions is 1.0–2.0% manganese, 0.2–0.6% silicon, and 0.2–0.5% carbon. These steels and certain modifications to be mentioned later stand next to the plain carbon steels in total tonnage produced.

Medium manganese steels respond well to normalizing and liquid quenching treatments and are seldom used in the fully annealed state. The effect of increasing manganese content on normalized steels is illustrated by the properties of Nos. 16–24 of Table LI. Further improvements in yield points and impact toughness may be obtained by quenching and drawing.

V. — Vanadium has been mentioned as a grain refining and cleansing element when used in certain wrought S.A.E. steels. It is not par-

ticularly effective in itself in increasing strength and hardenability but is widely used to intensify the effect of other elements. Vanadium is sometimes used alone to refine the grain size and thereby increase the toughness of large carbon steel castings. (See Nos. 25–27.)

Mn-V. — The carbon-manganese steels have a tendency towards coarse grain size and microsegregation in large castings and show corresponding lack of toughness, especially in the raw and annealed states. This condition can be altered by the addition of small amounts of certain elements. Nos. 28, 29, and 30 show the marked improvement in impact toughness produced by 0.10% vanadium.

Mn-Ti. — Titanium has also been used to improve the properties of medium manganese steel. (See No. 31.) Treatment with titanium is relatively inexpensive compared with most other alloying elements. Titanium is a strong deoxidizer and its action is largely that of cleansing the steel, very small amounts being present in the final composition.

Mo, Mn-Mo. — Molybdenum, like vanadium, is most effective when used with other alloying elements. However, carbon-molybdenum steels such as Nos. 32, 33, and 34 are used for improved properties at elevated temperatures and for their good welding characteristics. Higher tensile properties are obtainable when used with higher manganese contents as in Nos. 35 to 38. The beneficial effect of molybdenum on the elevated temperature properties of steel is indicated by the fact that a high percentage of the heat resisting steel castings used for steam fittings contain molybdenum, usually under 0.50%.[44]

The mechanical properties obtained by quenching and drawing an alloy cast steel containing 0.30% carbon, 1.25% manganese, and 0.25% molybdenum, a composition similar to No. 38, have been reported by Gregg[33] (quenched in water from 1600° F. and drawn at 1200° F.): tensile strength 110,000 lb. per sq. in., yield point 88,000 lb. per sq. in., elongation in 2 inches 18%, reduction in area 45%, Brinell hardness 215, Izod impact 54 ft. lb.

Mn-Mo-V. — High strength castings with satisfactory toughness can be made using proper proportions of the three alloying elements, manganese, molybdenum, and vanadium, as in Nos. 39 and 40.

Cu. — The effect of copper in steel castings has been a contro-

versial subject because of the alleged tendency of copper to segregate and produce " hot tears " during solidification.[24, 28, 35, 36] However, in recent years a considerable tonnage of copper steel castings has been successfully produced for automobiles and heavier machinery. These steels are unusual in that compositions containing over about 0.70% copper respond to precipitation hardening after normalizing or quenching. The hardening phase is pure copper itself, which precipitates as submicroscopic particles upon reheating to approximately 900° F. An example of the increase in strength produced by this treatment is given in Nos. 43 and 44. Commercial copper steel castings may also be given ordinary quench hardening heat treatments rather than this special precipitation treatment; however, the precipitation treatment has the advantage of giving more uniform results throughout large sections because rather low cooling rates are sufficient to retain the copper in solid solution, whereupon uniform hardening depends only on thorough heating in the precipitation treatment.

Cu-Mn-Si. — The three inexpensive alloying elements, copper, manganese, and silicon, have been successfully combined in the low carbon cast steels Nos. 46 to 49. The high elastic ratio obtained by the precipitation heat treatment, No. 49, is accompanied by good ductility.

Cr, Cr-V, Cr-Mo. — As in wrought steels chromium increases strength, resistance to abrasion, and hardenability upon quenching, and by the further addition of nickel, vanadium, or molybdenum the ductility and toughness of steel castings may also be improved. The effect of vanadium on the impact toughness of a chromium cast steel is illustrated by Nos. 52 and 53. The chromium-molybdenum varieties such as Nos. 54 and 55 are especially suited where deep hardening and a high elastic ratio are desired. High strength and hardenability are also available in the more complicated combinations of Nos. 56, 57, and 58.

Ni. — The last element to be considered is nickel. It is used alone or in combination with nearly all the other elements in various commercial cast steels. The low carbon nickel steel, No. 59, has been used successfully for locomotive frames and other castings where its toughness and general reliability are desirable.

Mn-Ni. — In the manganese-nickel steels, Nos. 63 to 67, nickel

tends to inhibit grain growth and increase the strength and toughness over that obtained in the straight manganese steels. A single normalizing and drawing treatment is sufficient to develop the high mechanical properties required in castings for light weight railroad cars.[37]

Ni-V. — The nickel-vanadium alloys, Nos. 68 to 72, have been used for heavy castings. It has been found that a high combination of properties is obtained when the carbon content is under 0.30%. The low carbon grade, No. 68, develops properties similar to No. 59 and has been used for locomotive frame castings.

Ni-Mo. — The nickel-molybdenum steels, Nos. 73 to 75, unlike their wrought steel counterparts, have found little application, although it is evident that their mechanical properties are good.

Ni-Cr. — As in the wrought S.A.E. steels, nickel-chromium combinations offer excellent possibilities for high strength alloys, especially when liquid quenching treatments are applicable. The properties of several normalized nickel-chromium steels are given in Nos. 76 to 81. The latter alloy is typical of the high carbon abrasion resisting castings used for excavating buckets, tractor shoes, dies, rolls, crusher jaws, and heavy gears.

Ni-Cr-Mo. — Alloys Nos. 82 to 85 show the possibility of producing very high strength castings of good ductility by using a proper combination of alloying elements and a heat treatment consisting of a single normalize and draw.

High Carbon Graphitic Steels. — The preceding table of casting alloys by no means exhausts the list of steels now being cast commercially. One of the most interesting metallurgical developments of recent years has been that of graphitic steel castings now used by the Ford Motor Company for crankshafts, wheels, and other parts formerly made from wrought steel or cast iron.[45] While these products are known as steel castings, it will be necessary to defer their consideration until graphitization has been studied in Chapter XIII.

Hadfield Manganese Steel. — High manganese steel was one of the first alloy steels to be commercialized. Patented by Sir Robert Hadfield in 1882, it is unique in that it has not been replaced in its own field of application by a superior alloy. The customary composition limits are:

10.00–14.00% Mn
1.00–1.40% C
0.04–0.10% P
0.01–0.03% S
0.30–1.00% Si

When steels of this analysis are properly heat treated, they exhibit a remarkable resistance to heavy abrasion and shock. They are used with marked success in dredging, excavating, and crushing machinery; as railroad crossings, switch-points, and frogs; and for a wide variety of machine parts in the oil well, paper making, lumber, cement, glass, sand and gravel, steel, and mining industries. Most of these applications require castings; however, it is possible to hot work this steel by rolling or forging, and certain types of trackwork, small forgings, rods, sheets, and wire are fabricated from ingots.

The acid or basic electric furnace, basic open hearth, or converter processes are used in making high manganese steels. The basic electric furnace gives the smallest loss of manganese when high manganese scrap constitutes part of the charge. A cupola is often used for separate melting of the ferromanganese.

The composition generally used is 12.0–13.5% Mn, 1.20–1.35% C, and 0.30–0.80% Si. In the cast condition the alloy consists of a mixture of iron and manganese carbides, a hard martensitic constituent and other complex decomposition products of the high temperature solid solution, together with portions of untransformed austenite. The alloy is air hardening, and, due partly to the excess cementite present, it is brittle in the cast state. A most important part of Hadfield's work in developing high manganese steel was the discovery that heating to high temperatures followed by quenching produced a marked change in the mechanical properties from brittleness to extreme plasticity and toughness. This remarkable result obtained by quenching from within the range 1830°–1940° F. is caused by the retention of the homogeneous austenitic structure which existed at the high temperature. (A ternary equilibrium diagram for this alloy system has been reported by Wells and Walters.[46]) Both the manganese and the carbon are soluble in the gamma iron at the quenching temperature and the high manganese content retards the $\gamma \rightarrow \alpha$ transformation so effectively that water cooling prevents transformation even in rather large sections up to 5 inches.

The resulting product has the characteristics of austenitic steels, including great toughness, high strength, high ductility, and almost complete lack of magnetism.

Unfortunately manganese does not impart corrosion resisting qualities comparable to those of the austenitic chromium-nickel alloy steels. If such had been the case, stainless steels would probably have been introduced before the year 1900, and their cost would have been considerably less than that of the well known high chromium-nickel alloys. A high manganese content is used, however, in combination with chromium in one type of corrosion resisting steel. (See Chapter XIV.)

The austenitic characteristics such as ductility and toughness do not explain the wear resistance of this steel when used for rock crushing and other forms of heavy abrasion; in fact, most other austenitic compositions would be quite ineffective for such usage. The high carbon, high manganese austenite is, however, especially susceptible to decomposition or transformation to martensite upon the application of cold work, and it is this property which makes it serviceable for the many severe applications mentioned above. The surface hardness is often increased to over 500 Brinell to a depth of ¼ inch after a period of heavy service, compared with 180–200 Brinell in the original quenched state. It has been found that high manganese steels have longer life under service conditions involving heavy impact and high pressures such as ore and rock crushers than they have in other applications were the wear is purely abrasive, as in sand blast nozzles. It is believed that the latter type of service is ineffective in work-hardening the surface.

In heating for quenching care must be taken that all parts of the charge are thoroughly soaked at the high quenching temperature. This requires a long heating time because the thermal conductivity is low. Intricate castings of irregular section are likely to warp because of the high temperature and drastic quench in cold water. This condition is aggravated by the combination of low thermal conductivity, about 1/6 that of iron, and high thermal expansion which are characteristic of this steel. In some cases the stresses set up by unequal cooling during casting or quenching cause cracking. A certain amount of warpage can be corrected by cold straightening

which is readily carried out in the plastic condition produced by quenching.

The mechanical properties of typical heat treated castings have been reported by Hall [47, 48]:

Tensile strength, lb./sq. in.	118,000
Proportional limit, lb./sq. in.	42,900
Elongation in 2 in., %	44.1
Reduction of area, %	39.0
Endurance limit, lb./sq. in.	39,000
Brinell hardness	190

Rolled products, such as rails, have even higher tensile strength with up to 70% elongation in 2 inches. The low proportional limit of both castings and rolled shapes makes it necessary to use special designs for rails and other sections so that the working stresses will not cause too great permanent deformation or distortion below the work-hardened regions.

It is not apparent from the mechanical properties given above that this steel would offer machining difficulties; nevertheless, it is not machinable on an economic basis. Grinding is much more satisfactory than machining where fabrication is necessary. The machining difficulties of high manganese steel are associated with the inherent toughness of its austenitic structure and its capacity for work-hardening by formation of martensite.

The non-magnetic nature of quenched high manganese steel is an important factor in its use as cover plates for electric lifting magnets and for other electrical applications. A non-magnetic facing of manganese steel on lifting magnets insures proper release of the load, and its shock and wear resistance are also useful properties for this type of service.

It is possible to cut high manganese steel with a gas torch and to make fusion welds by both gas and electric arc methods. Welding of worn or cracked parts was extremely difficult until special welding rods were introduced containing 3 to 5% nickel and 0.6 to 0.8% carbon, in addition to the usual high manganese content. One of the difficulties formerly encountered was precipitation of carbides and general hardening of the structure immediately adjacent to the weld. This is overcome by the stabilizing effect of nickel and the reduction

of carbon content in the fusion zone when the special rods are used. In gas welding the affected zone is more extensive than in arc welding, and heat treatment of the entire piece is usually necessary to restore the austenitic structure.   Using the improved welding procedure, extensive repairs may be made by building up the worn parts.

One of the limitations in the application of high manganese steel is its tendency to become brittle after heating to temperatures over about 700° F.   The cause is precipitation of carbides, the same difficulty mentioned above in connection with welding.   Rolled and forged products, including rods, sheets, and plate, are now made of the special nickel-manganese composition used for welding rods because it has better hot working properties and is less easily damaged by heating to moderately elevated temperatures.[48]   It also permits air cooling instead of quenching of light sections because of the greater stability of its austenitic structure.   The capacity of this material for work-hardening is illustrated by the properties of 0.018 inch wires (from a publication of The International Nickel Company) :

Hard drawn . . . .   303,600 lb. per sq. in. tensile strength
                                                    4.1% elongation in 8 inches
Annealed dead soft . . . . . .   133,000 lb. per sq. in. tensile strength
                                                    39.4% elongation in 8 inches

Considering the nature of the test specimen the ductility of the annealed wire is very high.

High Alloy Steel Castings. — Hadfield manganese steel may be classified as a high alloy steel while all the compositions of Table LI are medium alloy, low alloy, or carbon steel types.   Nearly all the remaining high alloy steel castings are iron-chromium alloys, modified by additions of nickel, molybdenum, silicon, tungsten, and a few other elements added in small amounts.   The development and application of these alloys in the fields of heat and corrosion resistance has closely paralleled that of similar wrought alloys.   This general subject will be studied in Chapter XIV.

REFERENCES

1. Rogers, G. P., " Statistical Data on Steel Casting Production in the United States,"
     Proc. Am. Soc. for Testing Mats., **32**, Part II, 50 (1932)

2. Cone, E. F., "Production Data of the Steel Foundry Industry for 1937," Metals and Alloys, **9,** 178 (1938)

3. Sauveur, A., "The Metallography and Heat Treatment of Iron and Steel," McGraw-Hill Book Company, New York (1935)

4. Keshian, H. G., "Dendritic Steel," Trans. Am. Soc. for Steel Treating, **17,** 321 (1930)

5. Sauveur, A., and Reed, E. L., "Dendrites in Nickel Steel," Trans. Am. Soc. for Steel Treating, **19,** 89 (1931)

6. "Symposium — Porosity in Steel Castings," Trans. Am. Foundrymen's Assoc., **42,** 323 (1934)

7. Briggs, C. W., and Gezelius, R. A., Four Papers on the solidification and contraction in steel castings, Trans. Am. Foundrymen's Assoc., **41,** 385 (1933); **42,** 449 (1934); **43,** 274 (1935); **44,** 1 (1936)

8. Batty, G., "Controlled Directional Solidification," Trans. Am. Foundrymen's Assoc., **42,** 237 (1934)

9. Batty, G., "The Influence of Temperature Gradients in the Production of Steel Castings," Trans. Am. Foundrymen's Assoc., **43,** 75 (1935)

10. Rose, W., "Green Sand for Steel Castings," The Foundry, **63,** 24 (March 1935)

11. Cone, E. F., "The Story of the Ford Cast Crankshaft," Metals and Alloys, **6,** 259 (1935)

12. Dwyer, P., "Steel Castings are Produced in Molds Bonded with Cement," The Foundry, **63,** 22 (Dec. 1935) and continued in Jan., Feb., and March 1936

13. Lorenz, F. A., "Notes on the Design of Steel Castings," Proc. Am. Soc. for Testing Mats., **32,** Part II, 58 (1932)

14. Wheeler, K. V., "Foundry Factors Affecting Steel Casting Design," Trans. Am. Foundrymen's Assoc., **40,** 125 (1932)

15. Spraragen, W., and Claussen, G. E., "Welding Cast Steel — A Review of the Literature to November 1, 1936," Supplement to The Welding Journal, **16.** 2 (April 1937)

16. Underwood, C. M., and Ash, E. J., "X-Ray and Welding — The Foundryman's Aid to Quality Castings," Trans. Am. Foundrymen's Assoc., **43,** 481 (1935)

17. Sampson, J. M., "Fabrication of Cast and Rolled Steel," Trans. Am. Foundrymen's Assoc., **43,** 351 (1935)

18. Quinn, T. S., "Fusion Welding as Related to Steel Castings," Proc. Am. Soc. for Testing Mats., **32,** Part II, 269 (1932)

19. "Recommendations of the A.F.A. Committee on Chemical Classification of Foundry Steels," Trans. Am. Foundrymen's Assoc., **43,** 581 (1935)

20. "Composition and Heat Treatment of Carbon and Alloy Steel Castings," Metals Handbook, 959, Am. Soc. for Metals, Cleveland (1939)

21. Sims, C. E., "Heat Treatment for Grain Size in Cast Steel," Metal Progress, **26,** 22 (Sept. 1934)

22. Grotts, F., "Dynamic Properties of Steel Castings," Trans. Am. Foundrymen's Assoc., **41,** 538 (1933)

23. Knox, J. D., "Heat Treating Steel Castings," The Foundry, **63,** 22 (Feb. 1935)

24. Armstrong, T. N., "Properties of Some Cast Alloy Steels," Trans. Am. Soc. for Metals, **23,** 286 (1935)

25. "The Scaling of Steel Castings," Trans. Am. Foundrymen's Assoc., **44,** 229 (1936)

26. Briggs, C. W., and Gezelius, R. A., "The Effect of Mass Upon the Mechanical Properties of Cast Steel," Trans. Am. Soc. for Metals, **26,** 367 (1938)

27. Heaslett, R. C., " Correlating Properties with Service Requirements," The Foundry, **64,** 26 (Dec. 1936) ; **65,** 28 (Jan. 1937)

28. Lorig, C. H., and Williams, C. E., " Physical and Mechanical Properties of Well-Known Cast Steels," Proc. Am. Soc. for Testing Mats., **32,** Part II, 114 (1932)

29. Connarroe, A. N., " Developments in Cast Steels Meets Demands of Service," Steel, **96,** 24 (May 13, 1935)

30. Grotts, F., " Representative Properties of Cast Medium Pearlitic Steels," Proc. Am. Soc. for Testing Mats., **32,** Part II, 197 (1932)

31. Strauss, J., and Norris, G. L., " Vanadium Steels," Metals Handbook, 592, Am. Soc. for Metals, Cleveland (1939)

32. Comstock, G. F., " Titanium Improves Pearlitic Manganese Steel," Metals and Alloys, **8,** 148 (1937)

33. Gregg, A. W., " Production of Alloy Steel Castings," Trans. Am. Foundrymen's Assoc., **42,** 56 (1934)

34. Strauss, J., and Norris, G. L., " Excellent Properties Developed by Vanadium Alloy Steel Castings," Trans. Am. Foundrymen's Assoc., **41,** 221 (1933)

35. Lorig, C. H., " Copper as an Alloying Element in Steel," Metals Handbook, 554, Am. Soc. for Metals, Cleveland (1939)

36. Finlayson, A., " Low Alloy Steel Castings," Metals and Alloys, **8,** 239 (1937)

37. Armstrong, T. N., " Nickel Steel Castings in Railroad Rolling Stock," Metal Progress, **33,** 163 (1938)

38. Walls, F. J., and Armstrong, T. N., " Properties of Heat Treated Nickel Steel Castings," Metal Progress, **33,** 387 (1938)

39. Sheehan, W. M., " Steel Castings Replace Built-Up Constructions," Metal Progress, **33,** 151 (1938)

40. Briggs, C. W., and Gezelius, R. A., " Detecting Defects by Radiography, Using Gamma Rays," Trans. Am. Foundrymen's Assoc., **39,** 65 (1931)

41. Lippert, T. W., " Modeling in the Foundry," The Iron Age, **137,** 26 (June 18, 1936)

42. Herington, C. E., " Steel Castings for Stiff Service," The Foundry, **65,** 50 (April 1937)

43. Dwyer, P., " Pennsylvania Steel Foundry Produces Heavy Castings," The Foundry, **63,** 22 (Nov. 1935)

44. " Standard Specifications for Valves, Flanges and Fittings for Service at Temperatures from 750–1100° F." — A157–36, A.S.T.M. Standards, Part I, 354 (1936)

45. McCarroll, R. H., and McCloud, J. L., " Ford Alloy Castings," Metal Progress, **30,** 33 (Aug. 1936)

46. Wells, C., and Walters, F. M. Jr., " Alloys of Iron, Manganese and Carbon — Part VII — Influence of Carbon on Thirteen Per Cent Manganese Alloys," Trans. Am. Soc. for Steel Treating, **21,** 830 (1933)

47. Hall, J. H., " Austenitic Manganese Steel Castings," Proc. Am. Soc. for Testing Mats., **32,** Part II, 238 (1932)

48. Hall, J. H., " Austenitic Manganese Steels," Metals Handbook, 567, Am. Soc. for Metals, Cleveland (1939)

CHAPTER XII

WELDING

In this chapter the welding and cutting of steel will be treated briefly from the metallurgical standpoint. The metallurgical operations of melting, carburizing, oxidation, degasification, casting, hot working, and heat treatment are all involved in modern welding and cutting operations.

Flame Cutting. — Flame cutting, which was once used mainly for salvaging scrap, has become an important fabricating tool for contour cutting of machine and structural parts from plate stock.[1]

The cutting tip contains jets for burning a fuel gas mixture for heating purposes, and usually a single jet for the delivery of pure oxygen at high pressures to the heated area. The gas for heating the steel is usually an oxyacetylene mixture but other cheaper fuels such as propane or city gas may also be used. It is only necessary to heat the steel above its ignition temperature of about 1500° F. before the stream of pure oxygen is applied. Once oxidation of the iron to Fe_3O_4 begins, the heat of the oxidizing reaction itself plays a large part in the continuation of the melting and oxidation process. Approximately 30% of the molten metal is removed without actual oxidation by the mechanical washing action of the stream of gas and burnt metal. Very heavy sections may be cut with the blow torch. Close dimensional tolerances can be maintained using machine operated torches.

In cutting mild steels the flame cut surface does not become embrittled and it will usually be suitable for any subsequent machining or forming operations necessary.

Comparative hardness tests of surfaces cut by five common methods were made by Zimmerman[2], using a 0.24% carbon, one-half inch structural steel plate. A superficial type of hardness test (10 kg. Vickers-Brinell) was used to explore the hardness of the plate at right angles to the cut at intervals of 0.002 inch. The following

condensed tabulation of results shows the moderate influence of flame cutting on the hardness of mild steel.

TABLE LII. EFFECT OF METHOD OF CUTTING ON THE HARDNESS AT THE EDGE OF A STRUCTURAL STEEL PLATE [2]

Distance from Surface, Inches	Vickers-Brinell Hardness				
	Machined	Hand Flame Cut	Machine Flame Cut	Sheared	Friction Sawed
0.000	164	204	218	245	345
0.002	137	197	206	240	333
0.01	134	200	190	237	227
0.04	133	170	174	193	158
0.10	137	158	151	169	136
0.25	137	155	143	159	134
0.50	138	139	139	136	138

The flame cut surfaces were softer than the sheared surface at all depths and did not develop the high superficial quench hardness of the friction sawed surface.

Steels with over 0.30% carbon, and alloy steels, are subject to air hardening to a depth of approximately 1/32 to 3/32 inch upon flame cutting. This may result in either microscopic or visible cracking, poor machinability, brittleness in forming operations, and warping. In order to extend the economies of flame cutting to many important medium and low alloy structural grades, several systems of heat treatment have been developed to overcome these difficulties.

The severed edge may be softened by the application of heat treating flames either preceding or following the actual cutting torch. Flame softening by this method requires careful control of heating and cooling just as in normal furnace type heat treatments. The source of heat for preheating or postheating is a bank of oxyacetylene tips actuated by the same mechanism that moves the cutting torch. Time-temperature diagrams by Moss [3] show the heat treating conditions which can be applied to flame cut surfaces. The effect of postheating is shown in Fig. 176. The very rapid heating and cooling of the plain flame cutting operation is caused by the chilling effect of the cold metal adjacent to the cut. Both air and quench hardening steels may be softened by reheating or tempering

below the critical range after cutting, as shown by the continuous solid curve in Fig. 176. Quench hardening steels may be annealed by reheating over the critical temperature. If the cooling rate from the annealing temperature is too fast to produce the desired softness in certain alloy grades, a secondary tempering heat may be used. The reheating methods would not be applicable to steels which are so strongly air hardening that they develop checks before the tempering heat can be applied. In such cases preheating flames may be

Diagram of Time-Temperature Relations Imposed on Flame Cut Surfaces, Post Heated in Various Ways
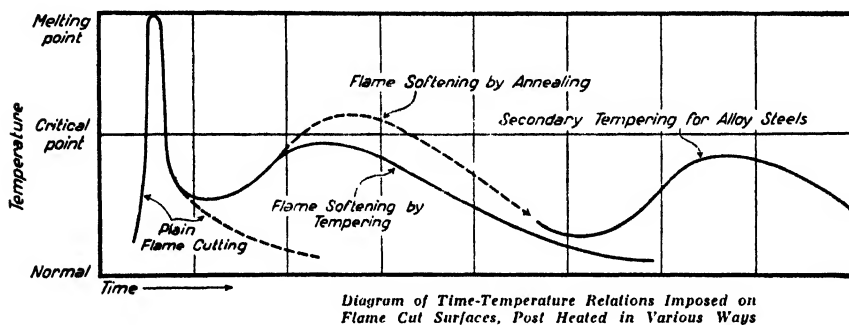
FIG. 176. — DIAGRAM OF TIME-TEMPERATURE RELATIONS FOR FLAME CUT SURFACES SHOWING THE APPLICATION OF POSTHEATING TEMPERATURES. (*Courtesy of Herbert H. Moss, Metal Progress.*[3])

applied so that the cooling rate from the cutting temperature will be reduced (not shown in Fig. 176). Heavy sections over about 1½ inches thick cannot be effectively reheated from the original plate surface or surfaces either before or after the cut. However it is possible to apply heat treating torches to the entire cross section of the newly cut surface as soon as the separation has been made.

Moss has given mechanical properties and other data for a number of structural steels after cutting and flame softening treatments.[3]

Pressure Welding. — Forge welding of the type used by the blacksmith before the era of fusion welding requires the application of sufficient heat to develop very high plasticity or a state of incipient fusion, followed by a considerable pressure applied by hand or power forging. The manner in which the actual union of the two masses is effected is still somewhat uncertain. It was once assumed

that a layer of metal actually melted at the interface when the hot metals were subjected to pressure, thus causing union upon freezing and subsequent hot working.

The recrystallization theory of the mechanism of forge welding does not assume partial melting. Two pieces of steel at a high temperature suitable for forge welding consist of crystals of gamma iron in a very plastic condition. The mechanical pressure applied is sufficient to cause at least partial breakdown of the films of gases or oxides which are normally present, thus bringing the surfaces into very intimate contact, and setting up a certain amount of strain within the crystals which will aid materially in bringing about rapid recrystallization. Under these conditions grain growth may progress across the original interface, thus effecting a perfect weld without melting, providing foreign substances can be removed. After the union is accomplished at high temperatures the continued hot working of the joint results in grain refinement and improved toughness. When the size and other physical conditions allow proper manipulation, and the weld is free from internal scale and dirt, very good results are obtainable by forge welding.

Resistance Welding. — The various forms of electric resistance welding also depend on high temperature, sufficient to cause melting or a state of very high plasticity, accompanied by a considerable unit pressure. Resistance welding is becoming one of the most important methods for joining metals.

Electric resistance welds may be applied to both butt and lap joints. Parts to be butt welded may be held together under pressure and heated by the passage of a very high current. Much smaller currents are sufficient in flash welding, a modification in which an arc is established before the pressure is applied. As the edges to be flash welded are brought together slowly, violent arcing volatilizes the metal at the first points of contact, tending to equalize the heat along the joint. After partial melting occurs the current is cut off and the parts forced together under high pressure, thereby squeezing out slag, oxides, and molten metal to form a flash around the joint. Wires are sometimes welded by another modification known as electropercussive welding.[4]

Spot and seam welds for lap joints are becoming increasingly

popular, especially in the fabrication of sheet metal products. Of the 3800 welds in the body and chassis of the 1936 Ford V8, 92% were spot welds.[5] Thousands of spot welds are sometimes applied to a single plate in the manufacture of duplex plates consisting of stainless steel, nickel, or other resistant materials welded to mild steel.

Spot welds have been called rivets without the holes. The success of the process depends to a large extent on the mechanical and electrical problems of timing, pressure application, and current regulation. The metallurgical problems include providing electrode materials of suitably high conductivity which will not fuse to the work or change in shape under the working pressures and the somewhat elevated temperatures of operation.[6] Water cooling is provided to prevent heating of the electrodes to very high temperatures. In addition to certain Cu-Be alloys (see page 148) other important types of electrode materials are essentially Cu-Cd, Cu-Cr, and Cu-W alloys, the copper content being over 95% in most cases.

The electrodes contact the work, usually two metal sheets, from opposite sides, making three metal to metal contacts in all; however, the resistance between the parts to be welded must be greater than that between either part and its electrode if the weld is to be successful. The electrodes and the work are a part of the low voltage circuit of a transformer. In a typical case in which 0.037 inch stainless steel sheets are spot welded, using $\frac{1}{4}$ inch diameter electrode tips and 250 lb. pressure, a current of 26,000 amperes is applied for one cycle or 1/60 second.[7] This very high current density of 400,000 amperes per square inch is necessary to cause fusion in such a short period of time. Current application for $\frac{1}{2}$ to 20 cycles is common in spot welding practice. The shortest possible time is usually desirable in order to confine the heated zone to the welded spot. This tends to keep the electrodes cool and prevent sticking, and to prevent discoloration of the outer surfaces of the work. In the case of stainless steels a low metal temperature will reduce the tendency towards precipitation of chromium carbides in the region adjacent to the weld. (See Chapter XIV.)

Aside from the metallurgical difficulty just mentioned, austenitic stainless steel is comparatively easy to spot weld. It has a high

electrical resistance which increases the heating rate, and its low thermal conductivity retards dissipation of the heat from the region of the weld. It has a rather wide plastic range before melting which is an advantage over certain other alloys which melt abruptly at the interface thus causing considerable squeezing in by the electrodes or, in case of poor timing, complete melting away at the electrodes.

Aluminum and its alloys are among the metals which are difficult to spot weld. In addition to their low melting point and low welding temperature range, they are handicapped by high thermal and electrical conductivity, and by the adherent and refractory oxide film which forms at ordinary and elevated temperatures.[8] An insulating film causes irregular passage of current and flashing, often accompanied by melting and porosity.

Although the heat is more nearly localized in spot welding than in any other form of welding, it is difficult to prevent some discoloration and denting-in of the outside surfaces. One surface may be left unmarked by using one large flat electrode with a dome shaped electrode of smaller contact area to localize the heat and pressure. Indirect spot welding and projection welding accomplish the same purpose.[9] In the latter process a projection is embossed or otherwise provided in one of the parts to act as a point of contact. Flat electrodes may then be used on both sides since the current will be concentrated at the projection.

Cross sections of spot welds in steel sheets are illustrated in Figs. 177 to 180. The zones which appear light in the photomacrographs and dark in the photomicrographs have been heated over the critical temperature range. The dendritic portions of these zones indicate that melting has occurred in the contact area.

The 10 kg. Vickers hardness impressions shown in Fig. 177 are evidence of the great increase in hardness of the heated and melted zones. The average hardnesses were: sheet 133 Vickers-Brinell (76 R " B "), heated zone 397 Vickers-Brinell (41 R " C "), and melted zone 388 Vickers-Brinell (40 R " C "). Referring to Fig. 122, page 218, it may be seen that the hardness of the welded metal checks the maximum hardness versus carbon content relationship quite well. It is clear that conditions in a spot weld are favorable
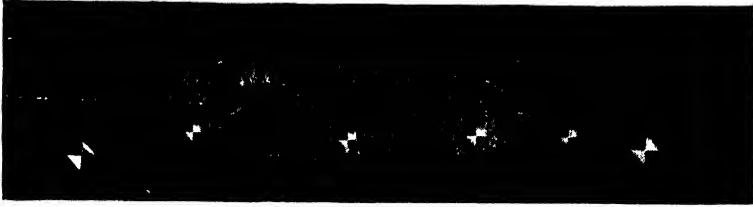
FIG. 177. — CROSS SECTION OF A SPOT WELD IN 0.14% CARBON, 0.42% MANGANESE STEEL SHEETS. (Etched with 3% nital.) Bottom electrode flat ($\frac{5}{8}$ inch diam.). Top electrode dome shaped. Voltage 4.4, time 10 cycles, force 1000 lb. X10.
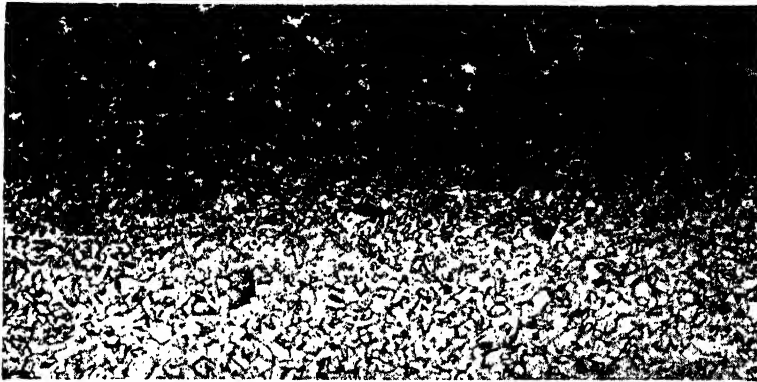
FIG. 178. — PHOTOMICROGRAPH AT THE RECTANGULAR AREA OF FIG. 177. (Etched with 3% nital.) X100.
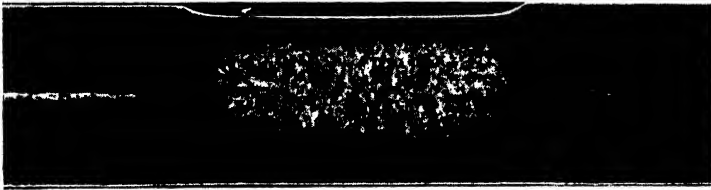
FIG. 179. — CROSS SECTION OF A SPOT WELD IN 0.03% CARBON, 0.28% MANGANESE STEEL SHEETS. (Etched with 3% nital.) X10.

to extremely rapid cooling, hence carbon and other elements which increase hardenability must be low to prevent brittleness.

A spot weld in a low carbon, cold reduced, box annealed, deep drawing sheet is shown in Figs. 179 and 180. The welding conditions were the same as for the higher carbon steel except the voltage was reduced to 4 volts and a slightly smaller electrode was used in

the top position.   In this case the hardness of the sheet was 78 Vickers-Brinell (36 R " B ") and that of the melted zone was 258 Vickers-Brinell ·(25 R " C ").   (Compare with Fig. 122.)

The transitions from the original sheet structures to the fused areas are shown in Figs. 178 and 180.   In Fig. 180 the carbide particles in A are transformed at a temperature just over Ac_1 to areas of high carbon austenite which become martensitic upon cooling, as in 180B.   Heating to a slightly higher temperature (as the fused zone is approached) causes transformation of a greater proportion of the metal to austenite.   The carbon content of the austenite is necessarily lower as its volume increases, and the transformation product obtained upon cooling has the appearance and properties of a low carbon martensite, as in 180C.   Heating to any temperature over Ac_3 produces a homogeneous austenite of very low carbon content which becomes a pseudo-martensitic structure of relatively high hardness as in Fig. 180D.   (This is a cast structure from the fused zone.)

The original pearlitic structure of Fig. 178 undergoes transformations in the same manner.

It is quite evident from these data that low carbon steels may be appreciably hardened provided very high cooling rates are used.

Seam welding is a modification of spot welding which may be used where pressure tightness is required.   The electrodes are two rotating copper or copper alloy discs between which the seam is passed. The current may be continuous or it may be interrupted at very short intervals in order to produce a series of overlapping spot welds.

Duplex Plate Joined by Welding. — An interesting application of pressure welding is a method used to make duplex plates, rods, tubes, etc.[10]   The surfaces to be joined are carefully cleaned and electroplated with iron, the thickness of the coating being about 0.006 inch.   The coated surfaces are pressed together and sealed by welding along the edges or by other means.   The assembly is then heated and either rolled or forged.   Usually one of the pieces is a high alloy tool or stainless steel and the alloying elements diffuse readily into the thin bonding layer of pure iron.

Fusion Welding. — The five foremost fusion welding processes are: (1) oxyacetylene, (2) metallic arc, (3) carbon arc, (4) atomic
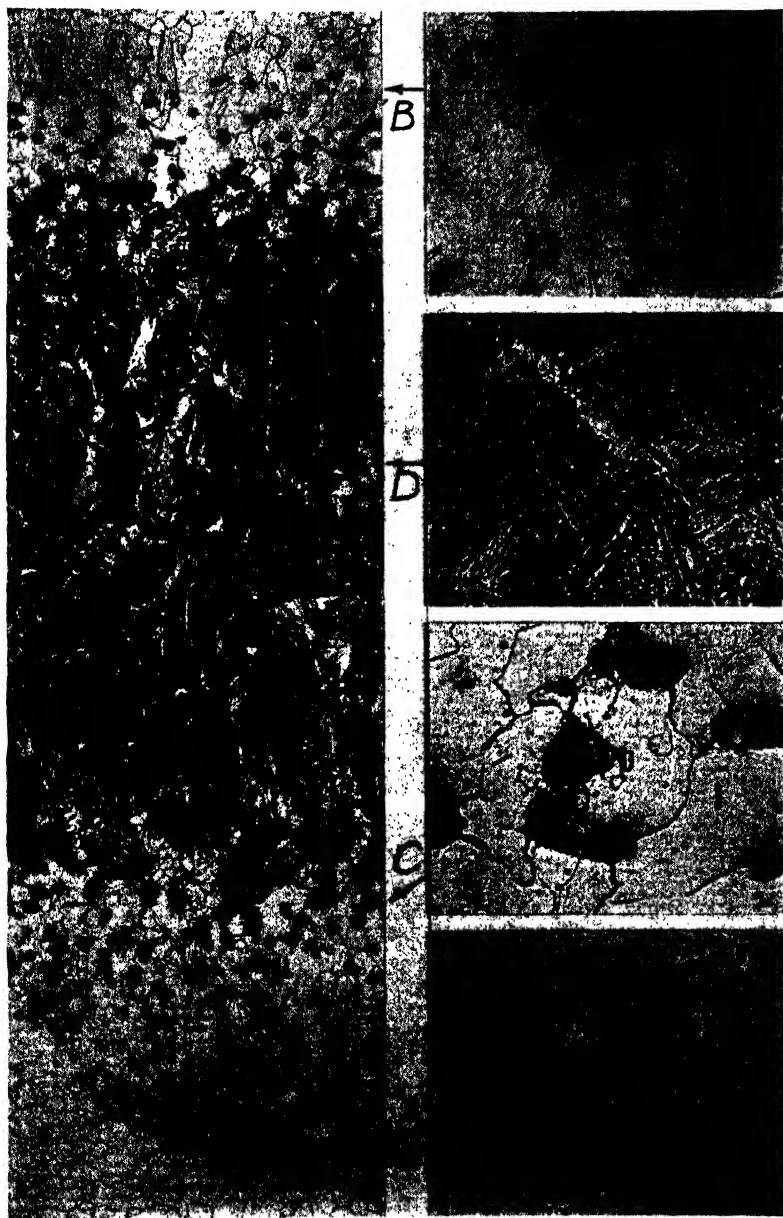
Fig. 180. — Photomicrographs of the Central Portion of Fig. 179. (Etched with 3% nital.) The panorama is at X100 — A, B, C, and D are at X500.

hydrogen, (5) thermit. In all cases a pool of molten metal is supplied at the joint and after solidification the bonding metal takes the form of a miniature casting. The greater part of the molten metal is usually supplied by a suitable filler rod; however, a portion of the parent stock is melted and mixed with the filler and in some cases no additional metal is needed. The intense heat required has an influence on the parent metal adjacent to the weld, producing zones of coarsening and of refinement. Forging or peening of fusion welds is sometimes possible but in general no mechanical pressure is used.

Fusion welding has become as indispensable as casting, machining, and heat treatment in the production of modern machines and structures. The rapid growth of welding in recent years is reflected in the ratio of steel welding wire to steel ingot production, which increased from 1.79 pounds of welding wire per ton of steel ingots in 1932 to 2.95 pounds of welding wire per ton of steel ingots in 1936. The metallic arc process is generally applied to automatic machine welding operations and to heavy work, hence arc welding rods make up by far the largest part of the total. The economic significance of these figures is further apparent when the cost of arc welding is considered. According to the following estimate by Smith[11], metal may be deposited from $\frac{1}{4}$ inch electrodes at a cost of about one dollar a pound.

TABLE LIII. COST OF DEPOSITING ONE POUND OF METAL
BY THE METALLIC ARC PROCESS

Electrode Size in Inches

	1/8	5/32	3/16	1/4	5/16	3/8
Labor	$1.150	$0.909	$0.758	$0.400	$0.280	$0.185
Overhead	1.150	.909	.758	.400	.280	.185
Power	.064	.059	.058	.055	.053	.051
Electrode	.150	.135	.127	.127	.127	.127
Cost of interruptions .	.050	.033	.022	.014	.008	.005
Total	$2.564	$2.045	$1.723	$0.996	$0.748	$0.553

Labor at $1.00 per hr.
Power at $0.02 per kw.-hr.
50% operating factor.
66⅔% deposition efficiency.

Courtesy of Edward W. P. Smith, Metal Progress.[11]

The competition between wrought steel fabricated by welding and steel and iron castings has already been mentioned. This competition is particularly keen in the production of frames for large electrical machines and machine tools. In many cases it is economical to flame cut, form, and weld steel plates costing nominally two cents a pound in competition with steel castings at nominally six cents a pound.[12] When multiple units are to be constructed, the use of castings is likely to be economical, while welded construction serves to best advantage when single units or a small number of units of special design are needed. The time and money spent for making patterns may exceed both the total construction time and fabrication cost for a single fabricated structure. Furthermore, changes are easier to make by flame cutting and welding of the fabricated structure, making it particularly suitable for the empirical development of new designs.

Oxyacetylene Welding. — In general, gas welding by oxyhydrogen and oxyacetylene methods is best adapted to the lighter types of work including sheets and tubes, to brazing, and to welding aluminum and other nonferrous metals. The oxyacetylene flame temperature of about 3300° C. (6000° F.) is not as high as that of the electric arc; consequently the heat cannot be localized as readily as with the latter medium.[13] In many instances this is considered to be an advantage since the structural transition of the affected parent metal is more gradual, and lower residual stresses are developed. Although the metallic arc seems to have certain inherent advantages in welding heavy sections, as in pressure vessels and power boilers, successful multipass methods have been developed using the oxyacetylene torch.[14]

The availability of both cutting and welding facilities and the compactness of the equipment are largely responsible for the universal use of oxyacetylene welding for general maintenance and repair service and for field operations such as pipe line welding.

Gas welding is particularly successful in the important field of seam welding of sheet metals.[15] The close control of the flame by selection of the proper size tip and by gas pressure adjustment is helpful in preventing thin sections from burning through. The fact that the speed of application of the filler rod is to some extent inde-

pendent of the flame used adds to the flexibility of this process. (This is also characteristic of the carbon arc.)

An advantage of oxyacetylene welding over some other types of fusion welding is the protection afforded by the flame against reaction between air and the heated metal. Neutral or slightly reducing atmospheres are easily obtained by adjusting the gas mixture. It has been observed that with a neutral flame very little carbon is burned out of the steel, and approximately 60 to 70% of the silicon and up to 50% of the manganese is lost. These losses are much lower than with unprotected arcs.[16] The flame used in the atomic hydrogen arc welding process also gives excellent protection.

The macrostructure of an oxyacetylene weld in a ¼ inch mild steel plate, Fig. 181, shows the extent of the fused zones and the grain coarsening in the plates adjacent to these zones. This weld had good ductility, taking a 180° bend without cracking. (See page 408.) The microstructure of the deposited metal,* Fig. 182, is that of a typical low carbon cast steel. It contains a small amount of well distributed pearlite.

A weld made under the same conditions except that the flame was slightly oxidizing instead of neutral was much lower in ductility and broke at 80° in the bend test. The deposited metal in this weld had a coarse ferritic structure with practically no pearlite. The lower ductility was caused, however, by the oxidized condition of the deposit rather than the lack of pearlite.

Bronze Welding or Brazing. — Weld deposits of manganese bronze and Tobin bronze based on the 60% copper-40% zinc alloy (see page 124) are generally made with the oxyacetylene torch. The parts to be joined are heated to a red heat but are not melted, hence the product is not strictly a fusion weld; however, the bronze generally penetrates into the base metal giving a good bond. The low temperature of the base metal is of special importance in brazing gray and malleable cast iron. Tensile strengths of 45–50,-000 lb. per sq. in. are obtained, which is generally quite satisfactory for welding castings. The deposit retains its plasticity to low temperatures, which reduces the residual cooling stresses. This is especially desirable in welding gray iron castings.

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* The structural changes in the parent metal adjacent to a fusion weld are illustrated on page 406.
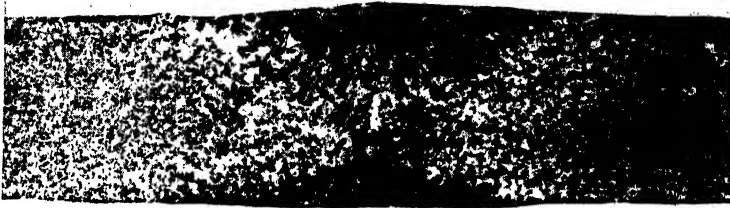
FIG. 181. — MACROSTRUCTURE OF A DOUBLE " V " OXYACETYLENE WELD. (Etched deeply with 5% nital.) X4. (*Weld furnished by Russell W. Pursel.*)

FIG. 182. — MICROSTRUCTURE OF DEPOSITED WELD METAL. (Etched with picral.) X100.

Arc Welding. — The speed of application, depth of penetration, and thoroughness of fusion which are characteristic of arc welding are made possible through the conversion of a large amount of electrical energy to heat in a short interval of time and in a closely confined space. For example, the arc produced at a ¼ inch steel electrode operating at 30 volts and 250 amperes draws 7.5 kilowatts or about 10 horsepower.[17] The high temperature produced in the arc, probably over 3700° C. (6700° F.), accounts for the rapid melting obtained.

There has been considerable interest in the nature of the arc, particularly the mechanism of transfer of metal from the electrode or filler rod to the deposit. In the case of bare electrodes and some coated types it has been shown that the metal passes intermittently in the form of large drops which may short circuit the arc instan-

taneously. In the case of electrodes heavily coated with a cellulose-sodium-silicate mixture there is evidence that the metal is transferred as a mist of very fine particles moving at high speed, and that less than one-tenth of one per cent of the arc space is filled with these particles.[18]

The general subjects of welding technique, welding equipment, power supply, use of alternating versus direct current, effect of polarity, and other operating details will not be discussed but are treated at length in many of the references given, including a 1600 page symposium on welding by the British Iron and Steel Institute [19] and " The Welding Encyclopedia ".[20] The " Welding Handbook " [21], recently published by the American Welding Society, gives full information on all phases of welding.

The three recognized arc welding procedures, metallic, carbon, and atomic hydrogen, will now be considered.

Metallic Arc Welding. — Bare electrodes were used in the early days of metallic arc welding. Although they are still used for certain purposes, they are rapidly being displaced by coated electrodes. The protective envelop of gases which covers the metal in the oxyacetylene process is absent in the bare wire metallic arc process, permitting the hot metal to react with the oxygen and nitrogen of the air. These reactions cause high losses of carbon, silicon, and manganese and the formation of oxides and nitrides which greatly reduce the ductility and toughness of the deposits. There are relatively few applications where welds of this type are now permissible, and the wide acceptance of arc welding as a means of fabricating pipe, structural work, machinery of all types, ships, boilers and other pressure vessels is the direct result of the development of electrode coatings.

The first coatings were very thin; in fact, the film of lime which often remains after wire-drawing was found to have a beneficial influence. Light flux coatings are now applied by dipping or dusting and consist of materials such as titanium and iron oxides, lime, soap, and similar substances which are readily volatilized below arc temperatures. These coatings furnish ionized vapors which tend to stabilize the arc and increase the melting rate, and in some cases to influence the depth of penetration. Although marked improve-

ments in these operating characteristics are made possible by the use of thin coatings, the ductility, fatigue strength, and impact toughness of the deposit are not greatly improved.

Shortly before the World War heavy coatings of slag-forming substances were applied to welding rods. Organic coatings which burned with a protective flame were also developed. Many of the best coatings now available are combinations of heavy slag forming ingredients and inflammable materials such as cellulose. The results obtained have been reported by a large number of operators and investigators, and in general the data are in good agreement.[18 to 25] Mathias[25] gives the following compositions of an electrode and the metal deposited when the electrode is used in both the bare and the heavily fluxed condition.

Per Cent	Electrode	Bare Electrode Deposit	Covered Electrode Deposit
Carbon	0.14–0.16	0.02–0.04	0.08–0.12
Manganese ..	0.40–0.50	0.25 max.	0.35–0.50
Silicon	Trace	Trace	0.05–0.30
Oxygen	0.02–0.06	0.15–0.25	0.06–0.15
Nitrogen	0.003	0.10–0.15	0.01–0.03

The welding rod is a typical low carbon rimming steel containing only a trace of silicon. The metal deposited from the bare electrode has lost most of its carbon content and has picked up a large amount of oxygen and nitrogen from the air. The increase in nitrogen content is particularly marked. Deposits with high nitrogen contents often contain visible nitride needles similar to those in Fig. 161.

The increase in silicon content of the covered electrode deposit is obtained from the coating. The limits for oxygen in this deposit indicate that conditions were somewhat oxidizing. In general this will depend upon the relative amounts of oxidizing substances such as iron oxide, and reducing agents such as ferrosilicon, which are contained in the coating. It is claimed that deposits from rods with cellulose type coatings contain oxygen and nitrogen contents on the minimum side of the above limits.[22]

Typical properties of metal deposited from bare and covered electrodes are compared with those of 0.15% carbon steel base metal in the following table by Hoyt.[22]

The low ductility, impact toughness, endurance limit, and density of the bare electrode deposit are directly associated with its high oxygen and nitrogen content. Nitrogen is particularly undesirable from the standpoint of embrittlement due to age hardening. Hansel and Larsen have shown that rapidly cooled weld deposits retain nitrogen in solid solution, which upon aging at room temperature or slightly higher precipitates as an iron nitride. Nitrogen contents under 0.05% are recommended to avoid embrittlement.[26]

TABLE LIV.   COMPARISON OF PROPERTIES OF METAL
DEPOSITED FROM BARE AND COVERED ELECTRODES

	Base Metal	Bare Electrode Deposit	Covered Electrode Deposit
Tensile strength, lb./sq. in.	53–57,000	50–60,000	60–70,000
Yield point, lb./sq. in.	28–32,000	38–45,000	40–50,000
Elongation in 2 in., %	35–40	8–12	25–40
Elongation, free bend %	—	10–20	40–70
Reduction of area, %	65–70	10–20	45–65
Endurance limit, lb./sq. in.	26–28,000	16–20,000	26–30,000
Izod impact, ft. lb.	75–80	5–15	40–50
Density, grams/cu. cm.	7.85	7.50–7.60	7.80–7.85

Courtesy of S. L. Hoyt, Trans. Am. Soc. for Metals.[22]

The above data show that the strength of the bare electrode deposit is satisfactory. However, it has been found that bare wire deposits in heavy sections such as one inch plates are low in strength because of their increasing unsoundness.[18]

The covered electrode metallic arc process has been particularly successful in making pressure vessels for the petroleum, chemical, and other industries. The average values of thousands of tests made on weld metal by five leading producers of pressure vessels have been reported as 56,500–71,400 lb. per sq. in. tensile strength; 22.5–38.6% elongation in 2 inches.[27]

The beneficial effects of heavily coated electrodes are the result of several different factors including the production of protective atmospheres or slags, the heat insulating effect of slag coatings, and the introduction of deoxidizers and alloying elements.

The protective action of the reducing flame produced by the burning of carbonaceous material has been mentioned. Some of the coating materials used for this purpose are paper, string, wood pulp, oatmeal, starch, tar, and synthetic resins. The principal products of dissociation and combustion are hydrogen and carbon monoxide.[18, 22] One advantage of coatings which consist mainly of combustible materials over those which depend mainly on slag formation is the adaptability of the former to vertical and overhead work.

There are a large number of slag producing substances in use including SiO_2, FeO, MnO_2, CaO, TiO_2, asbestos, hornblende, and feldspar. Titanium oxide (rutile) is particularly effective and large quantities are used in electrode coatings. In passing through the arc these coating materials help to shield the metal, increase the rate of deposition, and stabilize the arc. As a liquid coating over the molten bath the slag prevents reaction of the metal with the air after the shielding effect of the flame has been withdrawn. It also delays solidification, thus increasing the time available for evolution of dissolved gases and separation of entrapped slag particles. The composition of the slag must be such that it will be of proper density, surface tension, and viscosity to insure complete separation from the metal and formation of an effective cover. After both the slag and metal have solidified, the insulating effect of the former is still operative in reducing the cooling rate. This tends to reduce the magnitude of the cooling stresses and in the case of multiple pass welds it provides improved annealing conditions for the underlying passes.

Silicon, manganese, and titanium are often added to the coating in the form of ferroalloys. These and other elements may be added for their scavenging effect or for alloying purposes, or both. Alloys may also be introduced into the deposit by using alloy steel electrodes.

Sodium silicate, and to a lesser extent clay, is used as a binding material for heavy coatings.

Electrode coatings serve certain mechanical purposes such as insulating the rod, thus preventing arcing along the side walls when starting at the bottom of a weld in a heavy section. Ordinarily the wire tends to melt ahead of a heavy coating, forming a cup shaped

crater whose rim tends to confine and direct the deposition of metal and to reduce the loss of metal by spattering.

Unionmelt Welding. — A new electric welding process has been developed which requires special classification and consideration. A bare wire electrode is used and the welding heat is generated under a layer of slag. The powdered slag making materials are introduced at the weld by gravity through an overhead tube and hopper. They fuse into a very heavy coating over the weld deposit and around the electrode. It is claimed that no arc is formed and that the heat is produced by resistance to passage of current through the molten slag.

This process is limited to automatic machine welding in the horizontal position. Very high welding currents can be used, up to several thousand amperes, giving rapid melting and deep penetration. Sections over four inches in thickness may be welded in a single pass. The properties of these welds have been reported to be at least equal to those made by the best metallic arc practice. Although there are certain advantages in depositing the metal in several layers when making welds in heavy sections, the high ductility, soundness, and strength of Unionmelt deposits justify their use for certain classes of work. In shipbuilding, for example, this process has been applied to sections which can be fabricated in the shop under suitable conditions.[28] (The arc weld used for illustration in Figs. 183 and 184 was made by this process.)

In automatic welding machines a complicated mechanism is required to feed heavily coated electrodes continuously and to establish contact for conduction of the current. The use of bare wire in the above process is therefore an advantage. Other methods have been devised to wrap cellulose type materials around the electrode just before it enters the weld.

Carbon Arc Welding. — Arc welding with carbon electrodes is another process which is readily adapted to automatic machines. The carbon electrode burns very much slower than a metallic electrode, but the path of the flame is erratic unless superimposed magnetic fields are used to direct the arc.[29] In order to make deposits equal in quality to coated electrode metallic arc welds, the carbon arc must be shielded by an auxiliary gas flame, by burning an

impregnated " rope," or by supplying a heavy flux. A filler rod is used in some cases.

The carbon arc process is very fast and gives good penetration. It is generally used for machine welding applications where no filler is needed, as in edge welding and for unscarfed butt welds. It is also used for welding copper and other metals having high heat conductivity.

Atomic Hydrogen Arc Welding. — In the atomic hydrogen arc welding process the arc is struck between two inclined tungsten electrodes. A stream of hydrogen is passed through the arc where it is dissociated into atomic hydrogen, accompanied by a great increase in energy content. This energy is released as heat when the atoms recombine to form molecular hydrogen at the relatively cool surface to be welded. The concentration of heat is even greater than in metallic arc welding producing temperatures over 4000° C. (7200° F.).[30]

The hydrogen stream is very effective in preventing oxidation of the work, making it possible to weld many nonferrous metals and to deposit alloy steels of various compositions. Sound ductile deposits are obtained without using coated filler rods. The atomic hydrogen process is costly; therefore, it is restricted to special applications requiring high quality, and to alloy steels requiring special filler rods. Austenitic stainless steels are among the alloys which are successfully welded by this process.

A three phase three electrode atomic hydrogen arc process developed in France is said to give an extremely concentrated flame.

Thermit Welding. — Thermit welding utilizes the heat from the intensely exothermic reaction which takes place upon igniting a powdered mixture of metallic aluminum and iron oxide (Fe_3O_4) for the production of molten steel and a protective slag (Al_2O_3) and the development of sufficient heat to superheat the products of the reaction and fuse the sides of the metal to be joined.

The principal application of thermit welding is repairing large iron and steel castings and frames of heavy section. The surfaces to be welded are cut away to provide a suitable cavity, and a sand mold is built around the joint, then the mold and work are preheated. The thermit mixture is placed in a crucible above the mold and ig-

nited with a torch. The reduction of the iron oxide is very rapid and in less than a minute the hot steel can be introduced into the cavity.

The thermit method has been applied to certain types of repetitive work, notably the welding of rail ends.

Structural Characteristics of a Fusion Weld. — A cross section of a single pass arc weld is shown in Figs. 183 and 184. The structure of the mild steel plate, Fig. 184A, has been altered by heat treatment at successively higher temperatures up to its melting point. A part of the original stock has been melted and mixed with the filler rod, forming the cast structure of Fig. 184F.
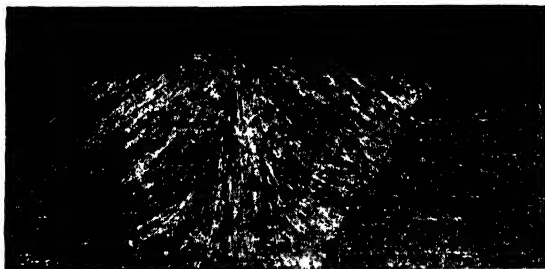
FIG. 183. — MACROSTRUCTURE OF A SINGLE PASS FUSION WELD MADE BY THE UNIONMELT PROCESS. (Etched by electrolysis in phosphoric acid.) X4.

The most notable changes in the parent metal are the marked grain refinement in Fig. 184B and C and the coarsening in D and E.

The major portion of B represents a zone which was heated to various temperatures within the critical range, Ac_1 to Ac_3. The individual pearlite areas are more finely dispersed than in the original plate but full grain refinement has not resulted. The fine grained structure of C indicates that this zone was heated above the Ac_3 transformation temperature. (The refined zone is relatively wide in this weld — one section of the panorama between C and D being removed.)

The structures in D and E were produced by heating over the austenite grain coarsening temperature, which appears to be rather well defined for this particular steel. Cooling is largely by conduction through the unheated parts of the plate and the rate approximates air cooling or normalizing. The angular, needle-like pattern of the coarsened regions is the Widmanstätten structure frequently observed in steel castings, overheated structures, and normalized steels.

These changes in the structure of the base metal are essentially the same in single pass fusion welds made by other processes.   When the heat is highly concentrated and melting is rapid, the zones are relatively close together.   On the other hand, most oxyacetylene welds have a larger affected area and a more gradual transition of structures.[31]

The mechanical properties of the various zones in the plate are not changed as markedly as might be expected from examination
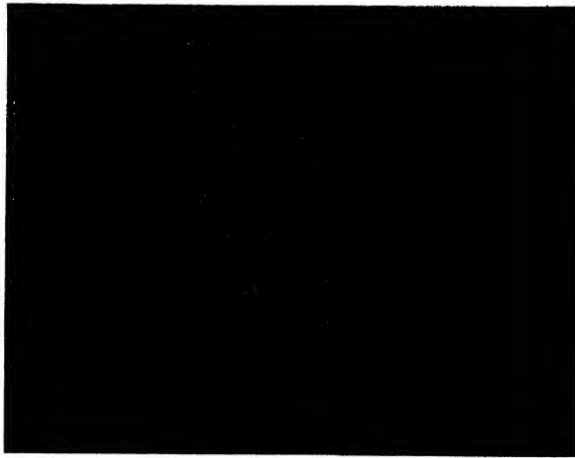
FIG. 185. — MULTIPASS METALLIC ARC WELD IN A $2\frac{3}{8}$ INCH PLATE.   (Etched.) X1.   (*Courtesy of J. C. Hodge, The Babcock and Wilcox Co.*)

of the microstructures.   This would not be true for higher carbon or alloy steels which are subject to air hardening.[32]   Each of the hardness values of the base metal, as given in Fig. 184, is the average of three 10 kilogram Vickers-Brinell tests made in regions having identical microstructures.   Nine hardness tests were made in the weld metal, giving a range of 146 to 155 Brinell.   The hardness of the weld deposit may be increased or decreased to balance strength and ductility requirements by changing the composition of the electrode or slag, or both.   The hardness of the coarsened zone is accompanied by a lower ductility than found in the unaffected plate.   However, the increased yield strength greatly reduces the amount of deformation taking place in this zone during a tensile test, so that

fracture is likely to occur farther back in the plate, unless unsoundness in the weld deposit itself causes failure.

Multipass or Layer Welding. — Welds are frequently built up by depositing several layers of metal in separate operations. In butt welding thick sections, especially by the metallic arc process, this method of deposition is very common. A typical arc weld in a heavy steel plate is shown in Fig. 185. Each deposit is cooled below the critical range before the next is applied; in fact, the surface is prepared by removing all slag and irregularities before another layer is deposited. The next pass tends to normalize or refine the grain size of a portion of the underlying metal by heating it through its transformation range. Carefully made welds of this type have excellent ductility and toughness.

It is sometimes considered desirable to apply two or more extra passes above the normal surface level. These may be machined off and the remaining metal at the surface will be refined.

The Free-Bend Test. — It is not the purpose of this book to describe methods of testing; however, an exception is made in the case of the free-bend test, which may not be as familiar to most readers as the tensile, hardness, and other frequently used tests.

In the development of welding technique the principal considerations are often general soundness and ductility, rather than tensile strength. Soundness may be judged by examining the fracture of a " nick-break " test specimen. The elongation obtained in a standard tensile test is quite satisfactory as a measure of ductility; however, it is often difficult or impossible to make standard tensile specimens from the deposited metal, consequently the free-bend test has been developed as a rapid method for checking the ductility.

The specimen used has a width equal to one and one-half times the thickness of the plate and may be of any length convenient to test. The welded section extends across the width of the specimen and the bead or beads are machined to the thickness of the plate. Gage lines are scribed in the weld metal across the width of the specimen at a distance of 1/16 inch within the lines of fusion with the base metal (as revealed by etching). In the case of a single " V " weld in ½ inch plate these marks will be on the top or working side, spaced approximately ½ inch apart. The ends of the specimen are bent slightly in

such a way that compression as a strut will cause extension on the gage-marked side. The compressive load need not be measured. When the first crack appears the load is released and the elongation between the gage marks determined. The per cent elongation is the free-bend ductility as reported, for example, in Table LV. The angle of bend is not usually reported unless the specimen makes a full 180° bend without rupture.

TABLE LV.   TENTATIVE SPECIFICATIONS FOR IRON AND STEEL
FILLER METAL — A.S.T.M. A205–37T [33]

| Grade | Deposited Metal | | Welded Joint | | Density |
	Tensile Strength lb./sq. in.	Elongation in 2 inches, % *	% of Base Metal Strength **	% Elongation in Free-Bend Test	
2	80 000	20	100	—	7.80
4	75 000	20	100	—	7.80
10	60 000	25	100	30	7.80
15	60 000	20	100	25	7.75
20	52 000	10	95	20	7.75
30	45 000	7	85	10	—
40	45 000	5	90	7	—

* These values apply after a low temperature stress relieving heat treatment has been applied. Lower elongations are specified for non-stress-relieved deposits.

** Minimum strength of base metal may be 55,000 lb. per sq. in. (A70 — firebox quality) or 60,000 lb. per sq. in. (A7 — for bridges).

The above tentative specifications of the American Society for Testing Materials (A205–37T) indicate in a general way the properties expected of modern fusion welds.[33] Welds in pressure vessels and other applications where maximum safety is imperative should conform to Grade 10. More detailed specifications and methods of test are often applied to specific types of work.

Residual Stresses in Welded Structures. — Internal stresses are developed due to restricted contraction upon cooling of welds, or more generally speaking these stresses are the result of the peculiar sequence of rapid expansions and contractions which are normally involved in making a welded joint. These stresses have been studied in considerable detail from an engineering viewpoint.[34] In many structures they are neglected, but in first quality welds for pres-

sure vessels, for example, adequate steps are taken to relieve internal stresses. It has been found that satisfactory results are obtained by heating to 1100°–1200° F., holding one hour per inch of thickness, and cooling slowly. Under certain conditions parts of a completed assembly may be treated separately, but wherever possible the entire unit is heat treated.

It is possible to refine the grain structure of the deposit and, to a large extent, to equalize that of the parent metal by heat treatment above the critical temperature range. However, the expense of this treatment is seldom justified.

Peening of Weld Deposits. — Arc welds may be either hot or cold worked, using light blows on a blunt nosed tool. An air hammer is used for production work. Results of tests show that when carefully applied, peening is capable of increasing the strength and reducing internal stresses, but usually the ductility and impact toughness are not appreciably improved; in fact, cold peening will lower the ductility.[35]

Hard Facing. — Facing of metal surfaces with hard wear resisting deposits is becoming an increasingly important application of the welding torch and electric arc.[23] The metals deposited range from inexpensive alloy irons and steels to high alloy steel, stellite, tungsten carbide alloys, and powdered mixtures of tungsten carbide and other hard carbide particles. (See Chapter XV.) Tools, rolls, and machine parts of many types are now hard faced, both in production and in building up worn surfaces.

REFERENCES

1. " Cutting of Metals (Gas and Electric Arc)," Metals Handbook, 921, Am. Soc. for Metals, Cleveland (1939)
2. Zimmerman, J. H., " Cutting Structural Steel," The Welding Journal, 15, 21 (Feb. 1936)
3. Moss, H. H., " Flame Softening of Cut Surfaces on Strong Steels," Metal Progress, 34, 14 (1938) — and — The Welding Journal, 17, 7 (Jan. 1938)
4. Hansen, K. L., " Digest of Welding Processes and Equipment," Metal Progress, 34, 115 (1938)
5. DiGiulio, A., " Metallurgical Problems of Welding in the Automobile Industry," The Welding Journal, 16, 8 (March 1937)
6. Larsen, E. I., " Fundamentals of Spot Welding with Particular Reference to Electrode Life," The Welding Journal, 15, 25 (April 1936)

7. Hutchins, W. C., " Resistance Welding Improved by Thyratron Control," The Welding Journal, **16**, 11 (Aug. 1937)

8. Hobrock, R. H., " Some Fundamentals of Spot Welding — Especially of the Light Alloys," Metals and Alloys, **6**, 19, 41, 67 (1935)

9. Roth, H. W., " Methods Used in Making Invisible Spot Welds," Steel, **101**, 44 (July 12, 1937)

10. Grimshaw, L. C., " New Method for Welding Together Ferrous Metals by Application of Heat and Pressure," Trans. Am. Inst. Mining and Met. Eng., **120**, 363 (1936)

11. Smith, E. W. P., " The Economics of Arc Welding," Metal Progress, **34**, 131 (1938)

12. Brinton, C. C., " Welding Large Structures," The Welding Engineer, 29 (Nov. 1937)

13. Slottman, G. V., " What Happens During Welding and Cutting? " The Welding Journal, **16**, 9 (Feb. 1937)

14. Plinke, G. W., " Gas Welding of Class I Pressure Vessels," The Welding Journal, **15**, 20 (April 1936)

15. Bainbridge, C. G., " Oxyacetylene Welding of Sheet Metals," The Welding Journal, **16**, 14 (April 1937)

16. Tichodeev, G. M., " Physical-Chemical Phenomena of the Transfer of Metal During Welding," The Welding Journal, **15**, 26 (March 1936)

17. Osborne, S. C., " What Happens in the Welding Arc? " The Welding Journal, **16**, 11 (June 1937)

18. Larson, L. J., " An Exploration of a Modern Welding Arc," The Welding Journal, **15**, 14 (Oct. 1936)

19. " Symposium on the Welding of Iron and Steel," Vols. I and II, The Iron and Steel Institute, London (1935)

20. " The Welding Encyclopedia — Ninth Edition," The Welding Engineer Publishing Company, Chicago (1938)

21. " Welding Handbook," Am. Welding Soc., New York (1938)

22. Hoyt, S. L., " Studies on the Metallurgy of Arc Deposited Weld Metal," Trans. Am. Soc. for Metals, **23**, 61 (1935)

23. " Welding of Metals," Metals Handbook, 220, Am. Soc. for Metals, Cleveland (1939)

24. Claussen, G. E., " Physical Properties of Welds," The Welding Journal, **16**, 16 (March 1937)

25. Mathias, D. L., " Fluxes — Their Function in Metallic Arc Welding," The Welding Journal, **16**, 20 (Feb. 1937)

26. Hensel, F. R., and Larsen, E. I., " Age Hardening Phenomena in Typical Fusion-Weld Metal," Trans. Am. Soc. for Steel Treating, **19**, 639 (1932)

27. Hopkins, R. K., " Construction of Welded Pressure Vessels for the Petroleum Industry," The Welding Journal, **16**, 16 (Jan. 1937)

28. Jackson, T. M., " Welding in Tanker Construction," The Welding Journal, **16**, 22 (Oct. 1937)

29. Rosenberg, E., " Apparatus for Automatic Arc Welding," Symposium on the Welding of Iron and Steel, The Iron and Steel Institute, London, **2**, 155 (1935)

30. Catlett, J. T., " Atomic Hydrogen Arc Welding," Metals and Alloys, **2**, 277 (1931)

31. Kinzel, A. B., " Metallurgical Effects Produced in Steel by Fusion Welding," Trans. Am. Inst. Mining and Met. Eng., **116**, 197 (1935)

32. Armstrong, T. N., " Arc Welding of High Carbon and Alloy Steels," Trans. Am. Soc. for Metals, **24,** 567 (1936)

33. " Tentative Specifications for Iron and Steel Filler Metal A205–37T," Proc. Am. Soc. for Testing Mats., **37,** Part I, 589 (1937)

34. Spraragen, W., and Claussen, G. E., " Shrinkage Stresses in Welding — A Review of the Literature to January 1, 1937," Supplement to the Journal of the Am. Welding Soc., **16,** 2 (Nov. 1937)

35. MacCutcheon, E. M. Jr., and Kingsley, D. M. Jr., " Peening and Its Effects on Arc Welds," Supplement to the Journal of the Am. Welding Soc., **16,** 22 (July 1937)

CHAPTER XIII

CAST IRON

The foundry industry consists of a large number of relatively small units, operating in every state of the union. Gray cast iron is the mainstay of the industry as shown by the following data from " The Foundry." [1]

FOUNDRIES IN THE UNITED STATES AND CANADA IN 1935

Gray Iron	3580
Malleable Iron	159
Steel	343
Electric Steel	227
Exclusive Nonferrous	1356
Nonferrous Departments of Iron and Steel Foundries	1504
Foundries Melting Aluminum . .	2394
Total Foundries	5550

Cast irons are probably the most interesting and diverse alloys to be studied, both from the standpoint of their metallographic structures and their engineering properties. Twenty years ago the bulk of the gray cast iron produced was of fairly definite chemical and structural composition. It was generally quite weak but could be cast easily and cheaply, was readily machined, and gave satisfactory wear resistance under many operating conditions. Certain high strength irons made in air furnaces and duplexed irons finished in electric furnaces were important exceptions. Ordinary gray cast iron served very well for frame castings for machines and machine tools, for pipe, ornamental castings, stove plate, and a great many other applications for which static and impact stresses are low. When applied to gears and other machine parts where service conditions are more severe, such irons were subject to excessive wear and breakage.

In the period following the World War the development of high

strength irons lagged behind the development of steel castings, forgings, and welding to such an extent that iron foundries lost an important part of their business. In more recent years the general improvement in cast iron practice has again put this material in a good competitive position and iron castings are being applied in increasing numbers to machine parts operating at moderately high stresses. The high strength and alloy gray irons and the pearlitic malleable irons now available combine many of the inherent advantages of ordinary cast irons with superior tensile properties.

FIG. 186. — WHITE CAST IRON STRUCTURE OF AN EARLY CHINESE CASTING. (Etched with 2% nital.) X100. (*Courtesy of M. L. Pinel, T. T. Read, and T. A. Wright, Trans. Am. Inst. Mining and Met. Eng.*[2])

History of Cast Iron. — Cast iron and wrought iron are the historical forebears of the modern iron and steel industry. Iron castings made in the early centuries of the Christian Era have been found in China and Japan. The Romans made cast iron statuary. Most of the existing specimens of ancient iron castings are statuary or ornamental panels; however, it is believed that early applications included ordnance pieces, stoves, chains, coins, and other useful articles.

Pinel, Read, and Wright [2] have recently published chemical data and photomicrographs of authentic Chinese castings made in the period 500–1100 A.D. Their photomicrograph of a cast iron panel taken from a pagoda built near Nanking in 1093 A.D. is reproduced here in Fig. 186. This panel was one of 1024 in the building. It has a typical hypoeutectic white cast iron structure. (See page 422.)

These early castings were probably made from iron reduced from iron ore by charcoal in small shaft type furnaces. The blast furnace began to assume its modern form in Europe in the fourteenth

century.[3]   Cannon and cannon balls were among the first products, and by the sixteenth century heavy ordnance was made in considerable quantities in England.[4]   Cast iron became available for general engineering and domestic purposes following the development of the cupola furnace in the eighteenth and nineteenth centuries.

The early cast irons became associated with tensile weakness and brittleness.   On the other hand, wrought iron of relatively great toughness and strength was already available long before cast iron was produced in large quantities.   Wrought iron could be produced at lower temperatures by reduction of the ore to a spongy mass of iron and slag without actually melting the iron.   However, the objects which could be made of wrought iron were very much restricted in size and shape because of mechanical limitations in working and welding large masses of hot iron.   Furthermore, cast iron was found to be superior to wrought iron for many purposes, especially for wearing surfaces and for frames for heavy machinery.   The early competitive situation between cast iron and wrought iron in the latter part of the eighteenth century is illustrated by an interesting study of anchors made for the British navy as reported by a famous English engineer in 1782.   (See Appendix, page 535.)

General Composition Range of Cast Irons. — In nearly all cases the ferrous alloys discussed in the preceding chapters were made in an open hearth, bessemer, or electric furnace from a charge consisting of pig iron, steel scrap, and slag forming materials.   During the refining period the carbon, phosphorus, manganese, and silicon contents of the pig iron were greatly reduced.   These elements amount to approximately 6 to 8% in pig iron and are reduced to 1.0% in a typical carburizing steel and to 0.1% in certain open hearth ingot irons.

Cast iron is also made from a blast furnace pig iron, but with relatively little change in chemical composition since the ordinary foundry cupola permits only minor readjustments of composition. When special irons are desired the nature of the charge may be altered by using steel scrap, special grades of pig iron, and ferro-alloys in the charge.   Additional elements may also be added at the runner or spout during tapping, in the ladle, or in a fore-hearth. Alloy cast irons and high strength irons may be melted in the cupola;

however, other furnaces capable of attaining higher temperatures and better metallurgical control are becoming increasingly important in this field. Although the charge of pig iron may be considerably diluted by the addition of steel scrap and then partially refined by special furnace practice, the product seldom contains less than about 5% of elements other than iron.

Cast irons cover a very wide range of compositions. The following limits include the most common varieties: 2.0–4.0% carbon, 0.5–3.0% silicon, 0.2–1.0% manganese, 0.05–0.80% phosphorus, 0.04–0.15% sulphur, 0.0–3.0% nickel, chromium, molybdenum, copper.

Graphite in Cast Iron. — The presence of large amounts of carbon and other elements in cast iron complicates the structure and imparts mechanical properties quite unlike those of steels. A large part of the total carbon present in gray cast iron will normally take the form of graphite flakes. As these flakes increase in size and amount, the tensile strength of the iron decreases and the fracture becomes darker in appearance. The density of flake graphite is about 2.25 which accounts for the fact that many cast irons have densities of 7.0 to 7.3 compared with 7.86 for pure iron. Upon examination of the photomicrographs which follow, it will be well to remember that 3% graphite by weight is equivalent to about 9.6% by volume.

Because of the important influence of graphite on the properties of irons it is desirable to be able to classify the type and amount of graphite observed in the microstructures. Mahin and Hamilton [5] have recently proposed the following classification of types based on the size and distribution of the graphite flakes.

No. 1 — Flakes 4 inches or more in length at X100
No. 2 —   ”   2 to 4 inches in length at X100
No. 3 —   ”   1 to 2   ”   ”   ”   ”   ”
No. 4 —   ” $\frac{1}{2}$ to 1   ”   ”   ”   ”   ”
No. 5 —   ” $\frac{1}{4}$ to $\frac{1}{2}$ inches in length at X100
No. 6 —   ” $\frac{1}{8}$ to $\frac{1}{4}$ ”   ”   ”   ”   ”
No. 7 —   ” $\frac{1}{16}$ to $\frac{1}{8}$ ”   ”   ”   ”   ”
No. 8 —   ” $\frac{1}{16}$ inch or less in length at X100

Type A — Random orientation of flakes uniformly distributed.
Type B — Random orientation of flakes grouped in rosettes.
Type C — Random orientation of flakes grouped in dendrites.
Type D — Dendritic orientation of flakes.

Structures representing several of these classifications are illustrated in Figs. 187 to 192. The coarse interlocking flakes of Figs. 187 and 188 and the fine network structure of Fig. 191 are associated with tensile weakness. Relatively fine but well distributed graphite flakes, as in Fig. 189, are necessary for good tensile strength. Although we are concerned at present only with the nature of the graphite, it should be noted that the matrix (not revealed in these unetched photomicrographs) also influences the properties of the iron.

Length classification alone is not entirely adequate since the total amount of graphite present and its distribution may vary considerably. Three different types of graphite formations, all having approximately the same length of flakes, are shown in Figs. 190 to 192. The tendency towards grouping of the flakes, as in Fig. 190, is characteristic of many cast iron structures. (See also Figs. 194 and 195.) The presence of graphite in finely divided form in the interstices of primary dendrites, as in Figs. 191 and 192, is common in the lower carbon (hypoeutectic) cast irons. The closely interlocked arrangement of fine graphite flakes illustrated in Fig. 191 has been referred to as a pseudo-eutectic structure by Bolton [6, 7] and others.

Solidification of Cast Iron. — The structures of Figs. 191 and 192 are definitely hypoeutectic in nature; i.e., the carbon content is less than the eutectic composition.

The ordinary iron-iron carbide diagram of Fig. 84, page 158, cannot properly be used to study the solidification of cast irons because they are modified by silicon contents up to about 3% and by smaller amounts of other elements. A section of the ternary iron-carbon-silicon diagram * taken at a constant silicon content of 2%,

* In sections of this type the composition limits of certain phases and microconstituents may not be represented in the plane of the paper, consequently the relative amounts of such phases or microconstituents cannot be determined graphically by the method applicable to binary systems.

FIG. 187

FIG. 188

FIG. 189

FIG. 190

FIG. 187. — GRAY CAST IRON CONTAINING 3.58% TOTAL C, 0.46% COMBINED C, 2.36% SI, 0.19% P. (Unetched.) Type 3A by Mahin and Hamilton's classification. Brinell 153. X100.

FIG. 188. — SAME IRON AS FIG. 187 BUT DIFFERENT FIELD. (Unetched.) Type 4A by Mahin and Hamilton's classification. X100.

FIG. 189. — HIGH TEST GRAY CAST IRON CONTAINING 2.88% TOTAL C, 0.66% COMBINED C, 2.24% SI, 0.22% P, 0.53% NI, 0.25% CR. (Tensile strength 44,000 lb. per sq. in.) (Unetched.) Type 5A. Brinell 221. X100.

reproduced as Fig. 193, is more accurate for ordinary cast irons.[8] Comparing this diagram with Fig. 84 it is apparent that silicon shifts the eutectic and eutectoid compositions to lower carbon contents. Nevertheless, the irons represented in Figs. 191 and 192 lie to the left of composition " C " and the white areas in these un-
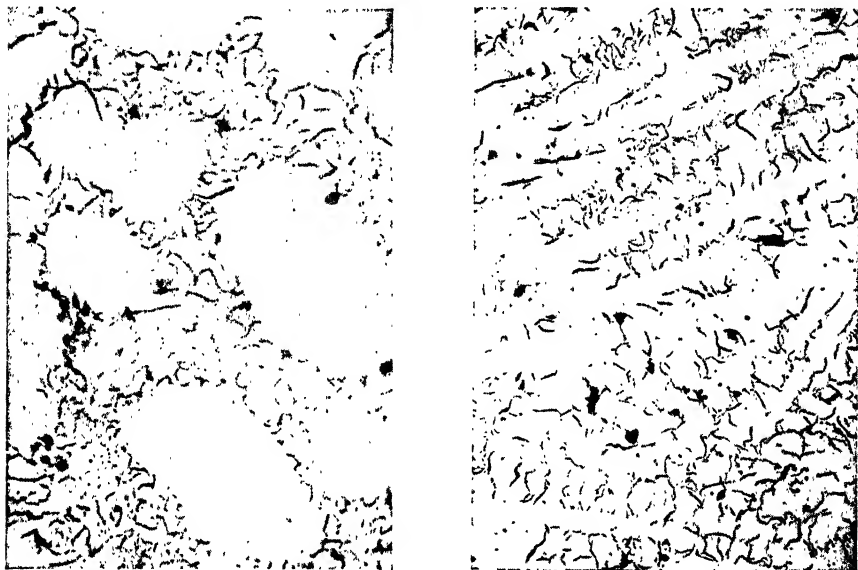
Left. FIG. 191. — GRAY CAST IRON. (Unetched.) Type 6C. Brinell 183. X100.
Right. FIG. 192. — STRUCTURE NEAR THE SURFACE OF A HIGH STRENGTH GRAY IRON CASTING CONTAINING 2.76% TOTAL C, 0.69% COMBINED C, 2.29% SI, 0.26% P, 0.44% NI, 0.20% CR. (Tensile strength 38,800 lb. per sq. in.) (Unetched.) Type 6D. Brinell 238 near surface, 222 at center. X100.

etched photomicrographs solidified as primary austenite dendrites in accordance with the diagram. As freezing progressed the carbon content of the melt increased until it reached the eutectic composition. The sequence of events thereafter is a controversial subject centered about the mechanism of graphite formation.

It was formerly assumed that a part of the graphite originally present in the pig iron remained undissolved and suspended in the molten iron at the temperatures ordinarily attained in the cupola. Such particles would act as nuclei for the crystallization of coarse graphite flakes upon solidification of the gray iron casting. The

principal evidence upon which this theory was based is the fact that superheating the iron in electric furnaces, or by other means, generally gives a much finer form of graphite, presumably through solution of carbon in the melt at high temperatures. However, both experimental and theoretical data indicate that free graphite cannot long remain in equilibrium with molten iron even at ordinary furnace temperatures, but should dissolve in the iron.
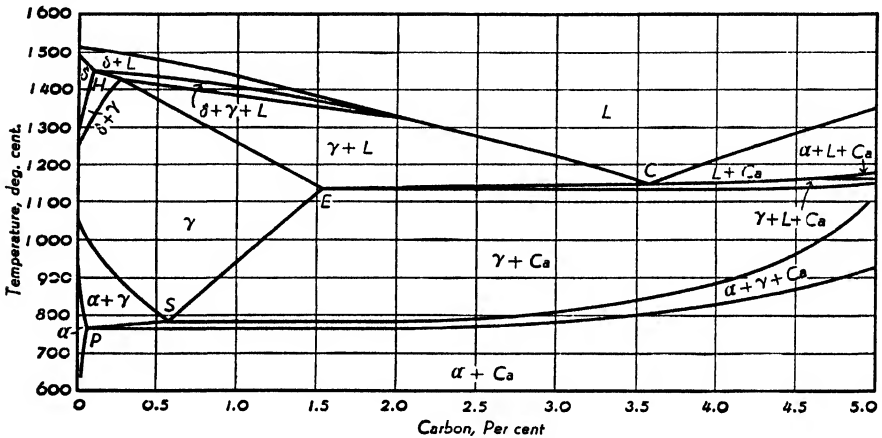
FIG. 193. — CONSTANT SILICON SECTION (2% SILICON) OF THE TERNARY IRON-CARBON-SILICON DIAGRAM. (*Courtesy of Greiner, Marsh, and Stoughton, Alloys of Iron and Silicon.*[8])

One of the alternative explanations for the effect of superheating is that the reduced viscosity of the iron at very high temperatures permits rapid separation of suspended silicates and other non-metallic inclusions which might otherwise act as nuclei for coarse crystallization of graphite. Molten iron in this clarified condition would be subject to supercooling on casting, and crystallization would be very rapid once it started, resulting in a fine structure. This is called the " silicate slime " hypothesis.

It is the opinion of many investigators that solidification of all normal and superheated irons begins by the separation of austenite, cementite, or the (δ + Fe$_3$C) eutectic in accordance with the equilibrium diagram. If this is the case, graphite formation results from the decomposition of solid Fe$_3$C according to the reaction:

$$Fe_3C \rightarrow 3Fe + C \text{ (graphite)},$$
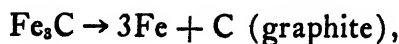
and the extent of graphitization depends upon the stability of Fe_3C at high temperatures.

Experiments on the effect of gas content upon graphitization and the general mechanism of solidification of cast irons have been reported by Boyles.[9]

Left. FIG. 194. — GRAY CAST IRON — TYPICAL " ROSETTES " NEAR THE SURFACE OF THE CASTING. (Unetched.) X100.

Right. FIG. 195. — SAME SAMPLE AS FIG. 194 — " WHORLS " AT THE CENTER OF THE CASTING. (Unetched.) X100.

Bolton [7] has also presented data on graphitization in which the importance of cellular and dendritic structures in hypoeutectic cast irons is emphasized. Bolton favors the hypothesis that graphite forms from cementite soon after the solidification of hypoeutectic or eutectic irons, and that the resulting graphite areas are definitely related to the previously existing eutectic (ledeburite) or primary cementite areas.

The forms which the graphite flakes may take are shown by Bolton to be related to the cooling rate during solidification. For example, Figs. 194 and 195 contain " rosettes " and " whorls," respectively, according to his classification. The fine graphitic structure in the centers of the rosettes of Fig. 194 is a pseudo-eutectic, as in

Fig. 191, and since the formation of this structure is favored by rapid cooling, it may be found in the surface layers of a casting in which the remainder of the structure consists of coarser graphite formations as in Fig. 195.

It is generally agreed that hypereutectic cast irons containing more than about 3.5% total carbon (depending on the silicon and phosphorus content) are likely to contain very coarse graphite flakes which begin to form directly from the melt or by decomposition of primary cementite very soon after it solidifies from the melt. This form of graphite is generally avoided in commercial irons because of its very harmful effect on tensile strength.

FIG. 196. — FRACTURE OF AN IRON CAST AGAINST A CHILL PLATE. X1.

Considering only the occurrence and form of the graphite, it is obvious that the solidification of cast iron is a very complicated subject. Further details may be obtained in references 7 to 13.

Combined Carbon in Cast Iron. — In normal gray cast irons approximately 1.8 to 3.4% carbon appears as graphite. Upon cooling in the mold a relatively small amount of carbon, usually less than 0.7%, remains dissolved in the austenitic matrix at a temperature just above the critical transformation range. Upon further cooling the austenitic matrix transforms to pearlite or a mixture of ferrite and pearlite just as in eutectoid or hypoeutectoid steels; in fact, the matrix of such an iron can conveniently be compared with the structure of a carbon steel whose carbon content is approximately equal to the non-graphitic, or combined carbon content of the iron. (To be discussed further in the next section.) Higher combined carbon contents will give stronger, harder cast irons providing the graphite content remains unchanged. It should be remembered, however, that the weakening effect of coarse or interlocking graphite flakes may overshadow any improvement of the matrix.

Under certain conditions all of the carbon may be retained in the combined form, giving a very hard iron whose fracture will have a characteristic brightness quite different from the usual dull gray to black appearance of a gray iron fracture. Because of this difference in fracture the graphite-free irons have been called white cast irons.

Irons of certain compositions are so constituted that they may be either gray or white in structure depending upon their cooling rate. If a heavy section of such an iron is cast against a chill plate of iron or steel, the chilled surface will be white for some distance within the casting. The transition zone between the white and gray portions will consist of a mixture of cementite, pearlite, and graphite known as mottled iron. The depth of chill, as determined by the appearance of the fracture of a standard chill bar, is often used in the foundry as a rapid test for control purposes.[14]

The fracture of a chilled iron, showing the three structural zones, is illustrated in Fig. 196. The

FIG. 197. — SAME IRON AS FIG. 189. (Etched with picral.) X100.

Vickers-Brinell hardness of this iron varied from 240 in the gray to 520 in the chilled or white zone.

Microstructures of Gray Cast Irons. — The structural significance of graphitic and combined carbon in cast irons is best understood by a study of typical photomicrographs. Although unetched structures such as those shown in Figs. 187 to 192 are useful in determining the amount and condition of the graphite, in order to reveal the structure of the matrix it is necessary to apply etching reagents.

In examining cast iron microstructures it is well to remember that the matrix is capable of assuming any of the structural forms observed in steels. Ordinarily, however, the structure of the matrix

of a cast iron will be similar to that of an annealed or normalized steel whose carbon content is somewhat greater than the *combined carbon* content of the iron. The microstructure of a low carbon high strength cast iron is illustrated in Fig. 197. Because of the presence of 2.24% silicon and smaller amounts of other elements,
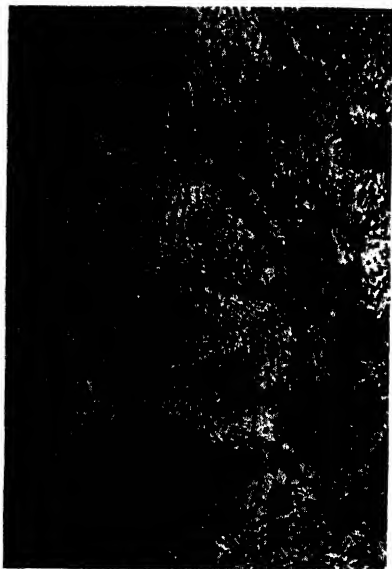
FIG. 198. — GRAY CAST IRON CONTAINING 3.4% TOTAL C, 0.67% COMBINED C, 1.9% Si, 0.21% P. (Tensile strength 36,100 lb. per sq. in.) (Etched with picral.) Brinell 182. X500.

its combined carbon content of 0.66% is sufficient to give a fully pearlitic matrix, whereas at least 0.8% combined carbon is required in pure iron-carbon alloys. The small white areas in Fig. 197 are not ferrite but steadite, a high phosphorus constituent. (See page 441.)

A pearlitic gray iron is shown at X500 in Fig. 198. The matrix of this iron may be compared with the pearlitic steel of Fig. 92. The hardness of the steel is 100 R " B " or about 240 Brinell compared with 182 Brinell for the iron. The tensile strength of the steel is about 110,000 lb. per sq. in. compared with 36,100 for the iron. The flake graphite is responsible for the lower hardness and strength of the cast iron, and even though the higher percentage of silicon and other elements in the matrix of the iron partially offsets the effect of the graphite, the net result is a very marked reduction in strength.

The structure of an iron having a hypoeutectoid matrix (approximately 0.5% combined carbon) is shown in Fig. 199. The white areas are ferrite and the dark are pearlite, as in hypoeutectoid steels.

Although the hardness of the hypoeutectoid iron of Fig. 199 is not appreciably different from that of the eutectoid iron of Fig. 198, when all other factors are held constant one would expect less hardness and strength in an iron having a hypoeutectoid matrix than in

one having a eutectoid matrix (a pearlitic iron). A soft, low tensile strength, gray iron having approximately 0.3% combined carbon is shown in Fig. 200. The lower hardness of this iron is, however, influenced by its coarser graphite as well as by its lower carbon matrix. (Note the magnification of Fig. 200 is only X100 compared with X500 for Figs. 198 and 199.)

Cast irons often appear to have a coarse cellular macrostructure which has been associated with chemical segregation during solidification.[15] In some cases it may be observed in the microstructure, as in Figs. 201 and 202, while in others it may be seen more distinctly with the naked eye upon examining fractures or polished samples. A method known as heat tinting is also very effective in revealing the macrostructure of cast irons.[9]

The light areas of Fig. 201 are a fine ferrite-graphite pseudoeutectic mixture. Pearlite and steadite are concentrated in the

FIG. 199. — GRAY CAST IRON. (Etched with picral.) Brinell 187. X500.

boundaries of these light areas. Mechanically this condition is undesirable from the standpoint of both strength and wear resistance. Vickers-Brinell (10 kg.) hardness impressions made in the light areas averaged 145 Brinell compared with 193 Brinell in the dark areas, and the tensile strengths of these microconstituents would probably show much greater divergence. The structure shown in Fig. 201 was observed only in the central portion of a small casting, the remainder of which had a normal microstructure with a slightly hypoeutectoid matrix. The occurrence of ferritic and pseudo-eutectic areas in small sand castings of normal compositions and in areas adjacent to the chill in chill castings indicates that certain critical cooling conditions favor their formation.[18]

Dendrites are often visible in both gray and white cast iron micro-structures.[9]   It is apparent that the dendrites of Fig. 202 are quite independent of the cellular macrostructure. At higher mag-nification, Fig. 203, it is apparent that the primary dendritic branches are essentially ferrite, and that the balance of the struc-ture within the dark network of Fig. 202 is the ferrite-graphite pseudo-eutectic. As in the case of Figs. 191, 194, and 201 this fine interlocking graphite structure is not conducive to high strength.

A microstructure consisting of fine graphite flakes in the inter-stices of primary dendrites having a pearlitic structure is shown in Fig. 204. This is another example of departure from the more usual structures such as Figs. 197 to 200 which are generally considered to be normal for ordinary gray cast irons.

FIG. 200. — STRUCTURE OF A CAST IRON LATHE BED CAST ABOUT 1913. (Etched with picral.)   Brinell 153. X100.

In the foregoing photomicro-graphs examples have been given of various combinations and forms of graphitic and combined carbon. The latter appears in the matrix as pearlite in gray irons and as massive cementite and pearlite in white cast irons, to be discussed later.

Factors Influencing Graphitization. — Having observed the ef-fect of graphite and combined carbon on the microstructure of gray cast iron, the factors controlling graphitization will be presented, followed later by further consideration of the relationship between structure and mechanical properties.

The total carbon content is influential in determining the propor-tions of graphitic and combined carbon present in cast iron. For example, a 3.5 % carbon iron will normally contain a large propor-tion of graphitic carbon while a 2.5 % carbon iron will often contain
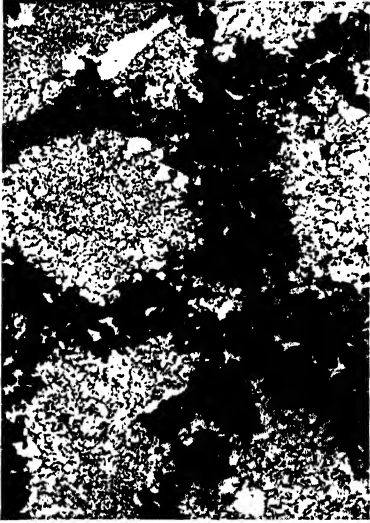
FIG. 201

FIG. 202
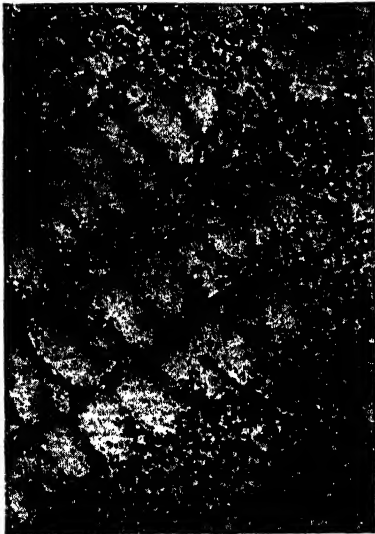
FIG. 203

FIG. 204

FIG. 201. — GRAY IRON CASTING HAVING A CELLULAR STRUCTURE CONTAINING PSEUDO-EUTECTIC AREAS. (Etched with picral.) X100.

FIG. 202. — CELLULAR STRUCTURE WITH PSEUDO-EUTECTIC AND DENDRITIC CHARACTERISTICS. (Etched with picral.) Brinell 193. X100.

FIG. 203. — SAME AS FIG. 202. X500.

FIG. 204. — DENDRITIC STRUCTURE WITH PEARLITIC MATRIX. (Etched with picral.) X500.

no graphite, thus high total carbon contents promote graphitization.

The rate at which a casting is cooled is an extremely important factor in determining the characteristics of the iron. A 3.5% carbon iron may be made practically white (free from graphitic carbon) if it is chilled sufficiently fast in the mold. On the other hand, a 2.5% carbon iron may be completely graphitized if it is cooled very slowly. Although we shall frequently consider the effect of section size, it should be remembered that section size is important largely because of its influence on cooling rate. Slow cooling promotes graphitization.

Silicon is present in cast iron as the principal element dissolved in the ferrite. It is possible, however, that a part of the silicon may be present in the cementite and even in the graphite. In any case silicon exerts a very strong influence in controlling graphitization. A 3.5% carbon iron, which would normally have a gray structure with 1.5% silicon present, would tend strongly towards a completely white structure with only 0.5% silicon present, thus high silicon contents promote graphitization. The combined effect of carbon and silicon is discussed on page 439.

The effect of sulphur depends upon the form in which it occurs. FeS has a strong tendency to stabilize Fe_3C and thereby reduce graphitization; however, there is usually sufficient manganese present in cast iron to insure MnS formation which alleviates the stabilizing effect of sulphur.

The effect of manganese is inherently tied up with that of sulphur. As in the case of steels, the MnS inclusions are much less detrimental mechanically than the low melting FeS network which forms when the ratio of manganese to sulphur is low. However, cast iron is not ordinarily hot worked, therefore its tenacity at high temperatures is somewhat less important. In the presence of a normal amount of sulphur, 0.05 to 0.16%, the principal effect of manganese is to increase graphitization by conversion of FeS to MnS. According to Spencer [14] the minimum chill (or the maximum tendency towards graphitization) occurs when the manganese content is 0.35% plus 1.7 times the sulphur content. Larger amounts of manganese tend to form Mn_3C, which reduces the tendency to-

wards graphitization. The net effect of manganese on the mechanical properties of cast iron is slight compared with the effect of silicon.

Phosphorus forms the hard constituent known as steadite but has a negligible effect on graphitization.

The influence of chromium and other alloying elements will be mentioned later.

White Cast Iron. — Several references have been made to the nongraphitic form of iron known as white cast iron. Hard irons are used for two types of castings. The greater part of the production is cast as white iron in ordinary sand molds and converted into malleable iron by heat treatment. The remainder is usually chill cast to develop a hard wear resistant surface with a gray iron interior.

Irons having a white structure throughout, especially those to be malleableized, are ordinarily cast in sand molds without the use of chills. In this case the chemical composition must be adjusted to prevent graphitization during casting. Usually the carbon content or the silicon content, or both, are relatively low. Special attention must be given to large castings to prevent the formation of flake graphite in the core.

Applications for chilled cast irons include freight car wheels, rolling mill rolls, and railway brake shoes. In their production, metal inserts in the mold serve to chill the surfaces which are to be hardened. If the composition of a cast iron car wheel is properly balanced, approximately one inch of hard nongraphitic surface metal can be obtained at the tread while the hub remains gray and readily machinable. Products of this type utilize the ability of the white iron structure to resist wear at high working pressures. The depth of chill is controlled by machinability requirements of the core and by the impact or shock resisting requirements in service. A large proportion of chilled metal reduces the impact resistance of the casting.

Other industrial uses of white or chilled castings include plowshares, liners and heads of crushers, balls and liners for ball mills, many types of wearing plates, stamp shoes, dies, and cams.

The high combined carbon content of white irons results in a

large proportion of massive cementite, the wear resistant constituent, as shown in Fig. 186.

Chill cast irons also contain massive cementite in the surface or chilled regions. The chilled structure shown in Fig. 205 (found in a thin web section of an ordinary gray iron casting) may be classified as a mottled iron structure because it contains graphite as well

FIG. 205. — CHILLED REGION OF A GRAY IRON CASTING. (Etched with picral.) Brinell 318. X500.

as massive cementite. A white or mottled structure such as this is very objectionable in a gray iron casting if it occurs in a region which is to be machined or if it must withstand shock in service.

It is obvious that the hardness and other properties of chilled irons will differ greatly between the white surface and the gray interior of the casting. For example, the coefficient of expansion of the chilled metal is about double that of gray iron; therefore, in order to prevent cracking, great care must be taken to apply heat uniformly and slowly in bringing chilled iron rolls up to working temperature in certain steel mill operations.

Whereas the cupola is by far the most frequently used medium for melting gray cast irons, the air furnace is widely used for melting the lower carbon white irons. The air furnace is a reverberatory type in which the metal is melted on a hearth. Although these furnaces are not regenerative, they retain many of the features of an acid open hearth furnace.[3] Since the metal is not in contact with coke as in a cupola, it is easier to keep the carbon content low. Absorption of sulphur is also reduced, compared with cupola melting, which reduces the tendency of the castings to check upon cooling, and has a favorable influence on graphitization during the malleableizing heat treatment.

TABLE LVI. TYPICAL COMPOSITIONS OF PLAIN CAST IRONS FOR COMMON APPLICATIONS

| Use | Composition, per cent | | | | | Typical Charge, per cent | | |
	C	Si	Mn	P	S	Pig	Return Scrap	Steel Scrap
Automobile cylinders	3.25	2.25	0.65	0.15	0.10	40	40	20
Automobile pistons	3.35	2.25	0.65	0.15	0.10	40	40	20
Automobile castings, general ...	3.40	2.60	0.65	0.30	0.10	50	45	5
Automobile piston rings — individually cast	3.50	2.90	0.65	0.50	0.06	60	40	—
Agricultural implements — medium sections	3.40	2.10	0.60	0.50	0.10	35	55	10
Agricultural implements — light sections	3.50	2.20	0.55	0.70	0.10	40	60	—
Machinery — sections not over 1 in.	3.25	2.25	0.50	0.50	0.09	50	50	—
Machinery — 1.5 in. sections	3.25	1.75	0.50	0.50	0.10	50	40	10
Machinery — 2 in. sections	3.25	1.25	0.50	0.50	0.10	50	25	25
Pressure castings — air cylinders ...	3.25	1.25	0.65	0.20	0.10	50	25	25
Gas-engine cylinders — light	3.40	1.75	0.80	0.35	0.09	55	20	25
Gas-engine cylinders — medium ...	3.40	1.50	0.80	0.35	0.09	55	20	25
Gas-engine cylinders — heavy	3.40	1.25	0.80	0.35	0.09	55	20	25
Valves and fittings	3.30	2.00	0.50	0.60	0.10	50	40	10
Firepots and kettles	3.50	1.15	0.80	0.10	0.07	70	30	—
Ingot molds	3.50	1.00	0.90	0.20	0.07	90	—	10
Pots for caustic soda	3.60	1.00	0.75	0.20	0.07	60	40	—
Light and medium sand-cast water pipe	3.60	1.75	0.50	0.80	0.08	70	25	5
Heavy sand-cast water pipe	3.40	1.40	0.50	0.80	0.08	60	25	15
Soil pipe	3.40	1.40	0.50	0.80	0.08	70	30	—
Wearing plates	3.20	1.10	0.85	0.35	0.17	*	43	30
Car wheels (0.90% combined C) ..	3.35	0.65	0.60	0.35	0.12	12.5	80	7.5
Chilled plow iron	3.60	1.25	0.55	0.40	0.10	45	40	15
Plow mold boards	3.75	0.85	0.50	0.40	0.10	45	40	15
Base for high-strength cupola iron; 5% ferroalloys to be added	2.75	2.25	0.80	0.10	0.09	—	10	85

25% scrap car wheels, 2% spiegeleisen.

From " Alloys of Iron and Carbon, Vol. II — Properties." [10] Originally published in Trans. Am. Foundrymen's Assoc. [19]

Other furnaces used for high temperature melting include the Brackelsberg [16], indirect arc [17], direct arc [18], and induction furnaces.

Typical Compositions of Gray Cast Irons. — The compositions listed in Table LVI [10, 19] include only ordinary cast irons (exclusive of high strength and alloy types) which have been found suitable for the given applications. In many cases there are several other compositions which might also be used.

The reader may at this point be able to associate some of the variations in the tabulated compositions, and the variations in properties which these suggest, with the service requirements or with the usual section size characteristic of a particular application. Further consideration of the properties of cast irons and the effects of the elements present should be helpful in interpreting their varied chemical compositions.

Mechanical Properties of Gray Cast Irons. — The mechanical properties of an annealed wrought steel are reasonably well defined if the chemical composition is known. The properties of cast irons are less easily related to their chemical compositions because of the marked influence of melting practice, mold practice, and section size, as well as heat treatment in some cases. Some of these variables have already been considered in a general way.

One of the principal differences between the properties of steels and ordinary grades of cast irons is the lack of ductility of the irons. The standard tensile test usually shows less than 1% elongation.

Other notable strength characteristics are illustrated in Bolton's stress-strain curve for a common grade of gray cast iron, reproduced in Fig. 206 (chemical composition: 3.37% total carbon, 0.34% combined carbon, 2.56% Si, 0.65% Mn, 0.30% P, 0.06% S, 0.14% Cu).[10] The curvature of the tensile stress-strain relationship, even at low stresses, makes it necessary to set an arbitrary standard for determining the modulus of elasticity. Thus at 12,000 lb. per sq. in. the effective modulus in tension is 12,000,000 lb. per sq. in., while at a stress equal to 25% of the breaking load, or 6700 lb. per sq. in., the effective modulus is 13,400,000 lb. per. sq. in. On this latter basis the moduli of elasticity for gray irons have been reported to be 10–20,000,000 lb. per sq. in.[20] compared with 28–30,000,000 lb. per sq. in. for steels. For cast irons having tensile strengths of 20–40,000 lb. per sq. in. the moduli of elasticity are within the narrower range of 14–18,000,000 lb. per sq. in. In

general, decreased size of graphite particles and amount of graphite present result in increased moduli and tensile strengths. These statements apply to irons in the as-cast condition — in some cases it is possible to raise the moduli by heat treatment.
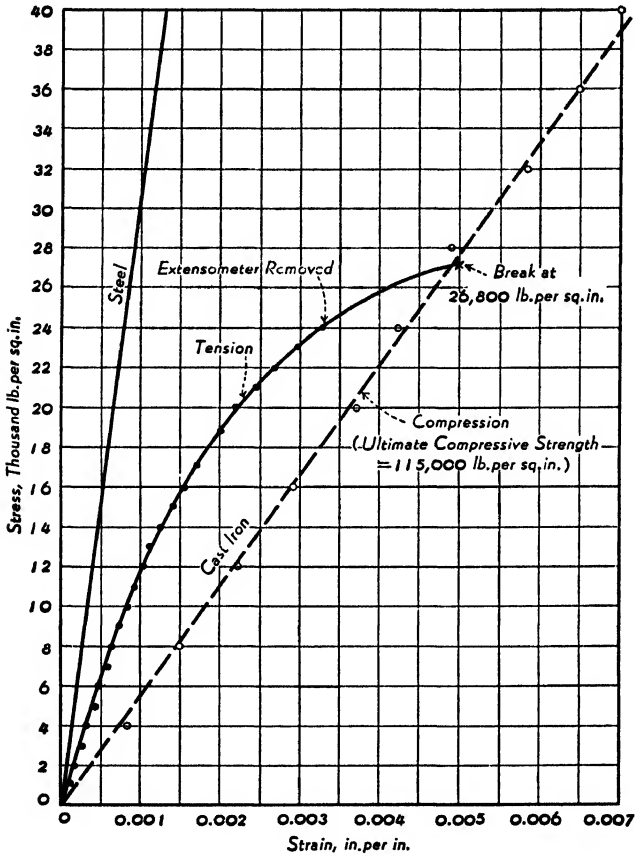
FIG. 206. — STRESS-STRAIN CURVES FOR A COMMON GRADE OF GRAY CAST IRON. (*Courtesy of J. W. Bolton, Alloys of Iron and Carbon, Vol. II — Properties.*[10]) Note: The line for steel has been added by the author.

Cast iron is unusual among metals in that it is very much stronger in compression than it is in tension, as illustrated by the following data for the iron whose stress-strain relationship was plotted in Fig. 206.[10]

Tensile strength, lb./sq. in. 26,800
Compressive strength, lb./sq. in. 115,000
Shear, on 0.505″ bar, lb./sq. in. 40,925
Brinell hardness 185
Deflection on 18″ span, inches 0.275
Transverse breaking load, lb. 2,310
Transverse modulus of rupture, lb./sq. in. . . . 61,200

The compressive breaking strength of this iron is 4.3 times its tensile strength. This ratio is somewhat higher than the following averages of data reported by an A.S.T.M. committee [20]: approximately 4 for iron whose tensile strength is 20,000 lb. per sq. in.; 3 for 40,000 lb. per sq. in.; and 2.5 for 60,000 lb. per sq. in.

Many applications of cast iron in frames and foundations depend upon its good compressive strength, low cost, ease of casting, and its ability to damp vibrations more effectively than an equivalent steel structure. The high damping capacity of cast iron, like many of its special properties, is associated with its graphite content.

The presence of innumerable structural discontinuities in the form of graphite flakes is believed to be responsible for the reduced notch sensitivity of cast iron compared with steel. Because of this characteristic, surface discontinuities in cast iron machine parts subjected to alternating stresses are less dangerous than in similar steel parts.

The Transverse Bend Test. — The transverse bend test is readily made in the foundry with simple equipment. A standard unmachined test bar is broken as a beam on supports spaced 12, 18, or 24 inches apart. The breaking load is a measure of the strength in bending and can be correlated with the tensile strength within the limits of variation of a given type of iron as produced by a given foundry. The maximum deflection at the center of the beam is often used as a measure of ductility.

Data for the three sizes of transverse test specimens and the A.S.T.M. grade classifications are given in Tables LVII and LVIII.

It should be remembered that the data in Table LVIII are minimum values for classification purposes. Nevertheless, these relationships are in line with actual test data and can be used to correlate the strength properties of different classes of irons.

TABLE LVII.   STANDARD TRANSVERSE TEST BARS
FOR CAST IRON

Test Bar	Nominal Diameter, Inches	Total Length, Inches	Distance Between Supports, Inches	Controlling * Section of Casting, Inches	$\frac{1r}{4I}$ (See page 436)
A	0.875	15	12	0.75	45.61
B	1.20	21	18	0.75–1.25	26.53
C	2.00	27	24	1.25–2.00	7.64
Old **	1.25	15	12	—	15.65

* Size of the important sections in the castings under consideration.
** This bar was widely used before the new standards were established. The breaking loads for this bar fall between those of B and C, and the deflections are lower than for any of the new bars, usually between 0.10 and 0.15 inch.

From A.S.T.M. Specifications A48–36.[21]

TABLE LVIII.   A.S.T.M. GRADE CLASSIFICATION FOR GRAY
IRON CASTINGS

Class No.	Minimum Tensile Strength lb./sq. in.	Minimum Transverse Breaking Load in Lb.			Test Bar B	
		Test Bar A	Test Bar B	Test Bar C	Modulus of Rupture lb./sq. in.	Mod. of Rupture divided by Tensile Str.
20	20 000	900	1800	6000	47 700	2.38
25	25 000	1025	2000	6800	53 000	2.12
30	30 000	1150	2200	7600	58 300	1.94
35	35 000	1275	2400	8300	63 600	1.82
40	40 000	1400	2600	9100	69 000	1.72
50	50 000	1675	3000	10300	79 500	1.59
60	60 000	1925	3400	—	90 100	1.50

From A.S.T.M. Specifications A48–36.[21]

The deflection requirements corresponding to these strength values are not specified by the A.S.T.M. but may be set by agreement between the foundry and the purchaser. In a group of 22 gray cast irons used for studying impact, transverse, tensile, and other properties, the deflection data for type B test bars covered the range 0.22 to 0.52 inches.[20]   Considering only those irons which contained over 3.0% carbon, the deflections were within the range 0.24 to 0.39 inches. A very high phosphorus iron and two high

alloy irons used in the original survey are not included in these figures.

The primary purpose of establishing three sizes of transverse test bars was to adapt the test results to the section size of the casting. (See Table LVII.) The necessity for this action becomes more apparent upon examination of Fig. 207, in which each curve represents change in strength with section size for a given class of iron.[22]

A value known as the modulus of rupture is often used in reporting the results of the transverse test. It is calculated from the formula:

$$\text{modulus of rupture} = \frac{Plr}{4I}$$

in which " P " is the breaking load in pounds, " l " is the span in inches, " r " is the radius of the bar in inches, and " I " is the moment of inertia of the bar in (inches)[4]. The modulus has the dimensions of a stress; i.e., lb. per sq. in.
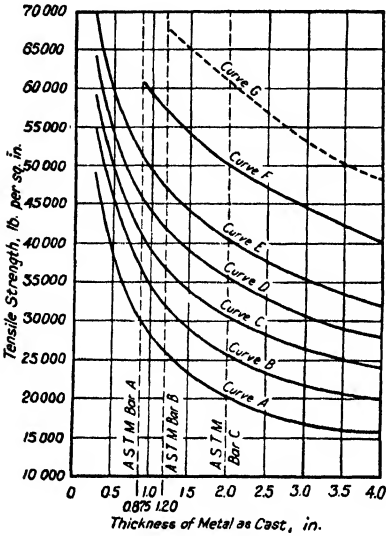
FIG. 207. — TENSILE STRENGTHS OF SEPARATELY CAST ROUND BARS OF CAST IRON. (*Courtesy of H. L. Campbell, Proc. Am. Soc. for Testing Mats.*[22])

If Hooke's law of stress-strain proportionality held up to the breaking point, the modulus of rupture would represent the maximum fiber stress in the bar at fracture. However, it has already been shown that tensile stress is not proportional to strain in cast iron, even at low loads, therefore the modulus of rupture is not a simply defined stress even though its formula has been taken from pure mechanics. The use of this quantity is justified, however, in that the values obtained are more nearly independent of the cross section and span used in the transverse test than are the breaking loads, P. For a given iron, both the modulus of rupture and the tensile strength decrease with increase in section size. The following data arranged by Bolton[6] from an A.S.T.M. report illustrate these points.

TABLE LIX.   TENSILE AND TRANSVERSE PROPERTIES OF THREE
GRAY CAST IRONS

Alloy	Total Carbon	% Si	% P	% Ni
I–AC	2.82	2.06	0.10	0.91
II–AC	3.40	1.15	0.21	—
III–AC	3.53	1.69	0.24	—

Alloy	Tensile Strength, lb./sq. in.			Transverse Breaking Load, lb.		
	A *	B	C	A	B	C
I–AC	65 750	58 900	53 450	1 875	3 175	10 925
II–AC	43 950	40 400	30 000	1 630	2 490	8 420
III–AC	38 600	31 600	24 730	1 380	2 230	7 245

Alloy	Modulus of Rupture, lb./sq. in.			Transverse Deflection, inches		
	A	B	C	A	B	C
I–AC	85 400	83 900	83 500	0.15	0.26	0.30
II–AC	67 700	66 000	64 300	0.16	0.27	0.39
III–AC	62 900	59 100	55 300	0.18	0.30	0.37

* Test bars A, B, and C.   (See Table LVII.)

It is generally the case that the skin of an iron casting differs from the interior to a somewhat greater extent than for steel and many nonferrous casting alloys.   The principal factors involved are pick-up of mold facing materials or sand, the chilling effect of the mold, and chemical segregation in the casting.   The net effect is usually an increase in the abrasive hardness of the skin.   It is interesting to note, however, that removal of the skin has been found to have little effect on the transverse modulus of rupture.[10]

It was observed in Table LVIII that at low tensile strengths the ratio of modulus of rupture to tensile strength is relatively high. The relatively high compressive strength of cast iron is undoubtedly effective in raising the transverse bending strength beyond the limits which would be predicted on the basis of tensile strength alone.   This condition is recognized when determining the allowable stresses in cast iron beams and pipes subjected to bending.   As the tensile strength approaches the compressive strength more closely in high strength irons, the superiority of the transverse strength over the tensile strength decreases.

Hardness and Other Properties. — The Brinell and other hardness tests have been very useful in controlling the properties of cast irons. An interesting diagram by Bolton and Bornstein showing the general relationships between hardness, strength, microstructure, machinability, and section size is reproduced in Fig. 208.[6, 10]
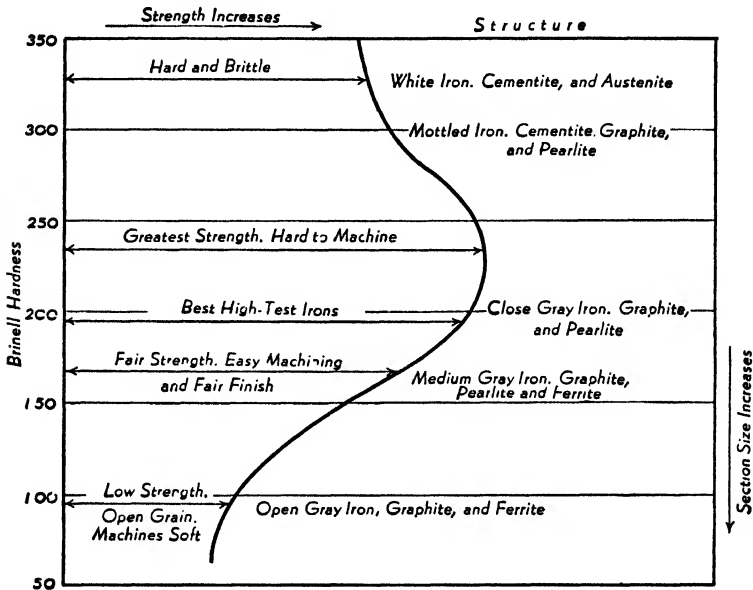
FIG. 208. — RELATION BETWEEN STRUCTURE, SECTION SIZE, AND PROPERTIES OF CAST IRON. (*Courtesy of J. W. Bolton and H. Bornstein, Alloys of Iron and Carbon, Vol. II — Properties.*[10])

The hardness of white irons may extend beyond the limits of this diagram to about 500 Brinell. (White irons usually consist of cementite and pearlite rather than cementite and austenite as given in the figure.)

The Brinell hardness of 18 ordinary irons reported by the American Society for Testing Materials [20] varied from 131 to 238 Brinell. The 4 high strength irons reported (over 40,000 lb. per sq. in. tensile strength) varied from 187 to 277 Brinell.

The 1933 " Symposium on Cast Iron " [23] and other references quoted contain detailed information on fatigue strength, impact toughness, wear resistance, machinability, growth at elevated temperatures, liquid fluidity, and many other properties of cast irons.

Structural Classification of Cast Irons. — Having reviewed the properties of irons in some detail and examined typical photomicrographs, the effects of individual elements on the structures and properties will be reconsidered and summarized using Table LX [24] and Fig. 209. [25]

The relationship between the carbon and the silicon contents is, in most cases, the controlling influence from the chemical standpoint.
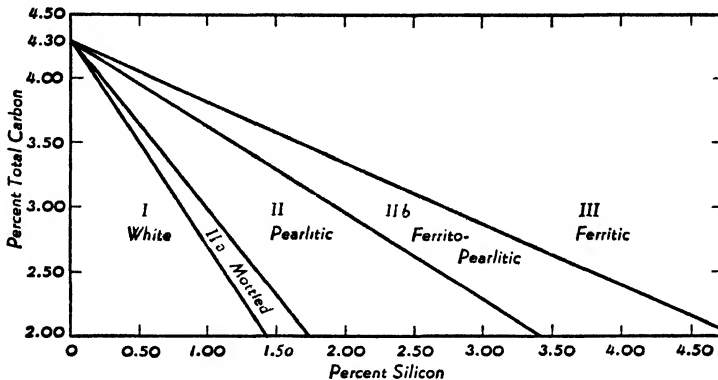
FIG. 209. — APPROXIMATE CONSTITUTIONAL DIAGRAM OF SAND CASTINGS OF TEST-BAR SIZE MELTED AND CAST AT NORMAL TEMPERATURES. [25] (*First proposed by E. Maurer.*)

Maurer's structural diagram, Fig. 209, which has been widely used to represent this relationship, is an approximation of the results to be expected with sand castings of test bar size melted and cast at normal temperatures, and although it does not strictly define the composition limits of the structures indicated, it is a convenient working basis.

Compositions falling in zone I of Fig. 209 are hard white irons free from graphite. (See Figs. 186 and 214.) Compositions in the transition zone, IIa, contain both graphite and massive cementite. (See Fig. 205.) Irons in zone II consist of graphite in a matrix of pearlite (Fig. 198). Irons in the transition zone IIb are similar to those in II but contain appreciable areas of ferrite in the matrix (Fig. 199). In zone III the matrix is predominately ferritic as in Fig. 200, and in extreme cases entirely ferritic. The size and amount of graphite ordinarily increases from zone IIa to III.

The modifying effect of cooling rate has been considered by Maurer and Holzhausser in a series of diagrams similar to Fig. 209.[8] In general, heavy sections or low cooling rates have the effect of shifting the lines to the left, whereas thin sections or chill casting shift the lines to the right.
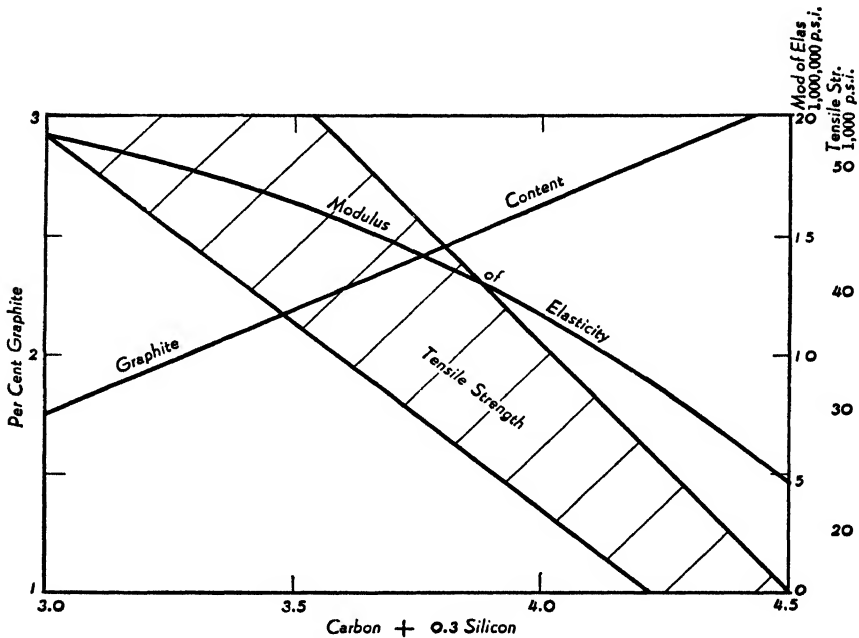
FIG. 210. — APPROXIMATE RELATIONSHIP BETWEEN GRAPHITE CONTENT, STRENGTH, AND ELASTICITY VERSUS THE QUANTITY: TOTAL CARBON PLUS 0.3 SILICON, AS DETERMINED FROM DATA ON GRAY CAST IRONS REPORTED BY AN A.S.T.M. COMMITTEE.[20] (*Courtesy of J. W. Bolton.* Replotted from the averages of data in Bolton's Figs. 190, 191, and 193.[6])

It is apparent from Maurer's diagram that carbon and silicon have a cumulative effect in promoting graphitization. Bolton has found that the amount of graphite present increases with the quantity: per cent total carbon plus 0.3 times the per cent silicon (C + 0.3 Si). The variation is approximately a straight line, as plotted in Fig. 210. Since the modulus of elasticity is determined to a large extent by the graphite content, it also is a function of the quantity C + 0.3 Si. The tensile strength is influenced by graphite flake size and distribution to a considerable degree, which may account for the

greater spread in the tensile strength data. Rather than give an average line for the tensile strength-composition function, the entire range of the A.S.T.M. data is included in the shaded area of Fig. 210.

If all of the silicon present is in solid solution in the ferrite, its marked ability to decompose Fe_3C is not easily explained. It is pos-sible that part of the silicon may dissolve in the cementite. Silicon also lowers the carbon content of the eutectic from the normal fig-ure of 4.3% carbon in pure iron-carbon alloys to 2.5% carbon at 5% silicon; thus the eutectic com-position of an ordinary gray iron containing 2% silicon is about 3.6% carbon, as indicated in Fig. 193. Most commercial irons have hypoeutectic structures despite their high silicon contents.

Effect of Phosphorus on the Matrix. — Phosphorus is present in cast iron in much larger amounts than in steels. Gray irons often contain 0.50% phosphorus com-pared with 0.05% in ordinary basic open hearth steels and

Fig. 211. — Steadite in Gray Cast Iron. (Etched with picral.) X1000.

0.005% in special open hearth grades. Phosphorus is readily de-tected as steadite in cast iron microstructures.

Because of its relatively high hardness, steadite tends to polish in relief; thus it is faintly visible in several of the photomicrographs of unetched specimens shown at the beginning of this chapter. Small white areas of steadite may also be observed in the etched photo-micrographs. A typical formation is shown at X1000 in Fig. 211. In many cases the internal markings are absent or less prominent than in this photomicrograph.

Steadite is essentially a eutectic of iron (containing phosphorus in solid solution) and Fe_3P. In gray irons its composition ap-

proaches that of the true binary eutectic; namely, 10.2% phosphorus and 89.8% iron. In white and mottled irons the high phosphorus constituent is believed to be a ternary eutectic containing Fe_3C as the third component.

Upon solidification of ordinary gray irons, the major portion of the alloy freezes at the eutectic temperature. The phosphorus present tends to segregate in the remaining melt which finally freezes as steadite, or the high phosphorus constituent, at about 1800° F. (The pure iron-phosphorus eutectic freezes at 1920° F.) As a result of this solidification mechanism, steadite is often found in the macroscopic grain boundaries, as illustrated in Figs. 201 and 202. The increased fluidity of high phosphorus irons has been attributed, at least in part, to the low freezing temperature of steadite.

In addition to imparting fluidity, the high phosphorus constituent increases the hardness of the iron because of its own inherent hardness. Phosphorus is generally kept under about 0.25% in automotive irons because higher amounts reduce the machinability at high cutting speeds.

High Strength Cast Iron. — High strength irons include a wide range of compositions, with and without alloy additions.

The principal feature of the non-alloyed type is the low carbon content, usually obtained by increasing the proportion of steel in the charge in cupola practice.[26, 27, 28]

High strength, low carbon, non-alloyed iron may also be obtained by melting a moderately low carbon charge in the absence of coke, as in electric furnace, Brackelsberg, or air furnace practice.

One of the advantages of the low carbon irons containing less than 3% carbon is their greater uniformity of properties when cast in various section sizes, compared with ordinary gray irons.

Both the low carbon irons and those containing 3.0 to 3.6% carbon are subject to improvement in strength and density by superheating to steel-making temperatures. (See page 420.) Before casting superheated irons, especially the low carbon variety, it is often necessary to inoculate the molten iron with graphite forming or graphite containing ladle additions in order to avoid the formation of low strength, cellular ferrite-graphite eutectic structures such as those shown in Figs. 191, 201, and 203. The principal

additions made to promote favorable graphitization are ferro-silicon, graphite, pig iron, or return scrap. It is believed that the deoxidizing effect of these additions may be responsible for the beneficial results obtained.

Meehanite irons are made by a patented process in which white or mottled basic compositions are graphitized by ladle additions of a calcium-silicon alloy. A wide range of irons having tensile strengths of 30–60,000 lb. per sq. in. may be produced by this method.

Regardless of the method of melting, a 50,000 lb. per sq. in. tensile strength iron will cost more to cast and to machine than a 30,000 lb. iron. Furthermore, it is likely to be less fluid and to have a greater shrinkage than the weaker iron, making the production of intricate castings difficult. There are innumerable applications where the lower strength irons are entirely adequate and should be specified.

Permanent Mold Castings. — In contrast with the low carbon type of high strength irons, a composition containing about 3.6% carbon and 2.5% silicon is being cast successfully in permanent metal molds. Although sand castings of this composition would have coarse, weak structures, the permanent mold castings have good strength and density. In addition to the effect of faster cooling, the pressure exerted by the mold during the period of graphite formation is said to be responsible for the fine grained structures.[29] Chilled iron castings of the high carbon low silicon type are also cast in permanent molds.

Centrifugal Castings. — Cast iron pipe is the principal iron product made by centrifugal casting.

A rotating water cooled metal mold is used in the deLavaud process. The composition of the iron is approximately 3.65% C, 1.80% Si, 0.45% Mn, 0.85% P, 0.075% S.[30] This composition would have but slight tendency to chill if cast in sand; however, in the water cooled metal molds up to 50% of the section is chilled. The excessive brittleness of the raw castings is relieved by annealing at 1700° F. After annealing the combined carbon content is reduced to 0.10%, indicating rather complete decomposition of the cementite to graphite and ferrite. However, because of the fineness of the cast structure and its high density, the tensile strength after heat

TABLE LX.   SUMMARY OF STRUCTURAL EFFECTS OF ALLOYS ON IRON

	Percentages Used in Pearlitic Irons	"Chill"	Effect on Carbides (at High Temperatures)	Effect on Graphite Structure	Effect on Combined Carbon in Pearlite	Effect on Matrix
Chill-inducing Elements						
Chromium..........	0.15–1.00	Increases[a]	Strongly stabilizes	Mildly refines	Increases	Refines pearlite and hardens
Vanadium..........	0.15–0.50	Increases	Strongly stabilizes	Refines	Increases	Refines pearlite and hardens
Mildly chill-inducing						
Manganese.........	0.30–1.25	Mildly increases	Stabilizes	Mildly refines	Increases	Refines pearlite and hardens
Molybdenum.......	0.30–1.00	Mildly increases	About neutral	Strongly refines	Mildly increases	Refines pearlite and strengthens
Mildly chill-restraining						
Copper.............	0.50–2.00	Mildly restrains	About neutral	About neutral	Mildly decreases	Hardens
Chill-restraining						
Carbon.............	—	Strongly restrains	Decreases stability	Coarsens	Strongly decreases	Produces ferrite and softens
Silicon.............	—	Strongly restrains	Decreases stability	Coarsens	Strongly decreases	Produces ferrite and softens
Aluminum..........	—	Strongly restrains	Decreases stability	Coarsens	Strongly decreases	Produces ferrite and softens
Nickel.............	0.10–3.00	Restrains[b]	Mildly decreases stability	Mildly refines	Mildly decreases and stabilizes at eutectoid	Refines pearlite and hardens
Titanium...........	0.05–0.10	Restrains	Decreases stability	Strongly refines[c]	Decreases	Produces ferrite and softens
Zirconium..........	0.10–0.30	Restrains	—	About neutral	—	Produces ferrite and softens

[a] Chill-inducing effect about balances chill-restraining effect of $1\frac{1}{2}$ parts of silicon or $2\frac{1}{2}$ parts of nickel.

[b] Chill-restraining effect about $\frac{1}{2}$ that of silicon.

[c] When added in small amounts and particularly when oxygen is also present.

Courtesy of Paul D. Merica, Trans. Am. Inst. Mining and Met. Eng.[24]

treatment is said to be about 32,000 lb. per sq. in., which is relatively high for an iron whose matrix is essentially ferritic.

A recent development in this process involves the use of a very thin layer of refractory material dusted on the mold wall just before pouring. The insulating effect of this film is great enough to prevent chill, giving a dense gray structure throughout the section.

A sand lined mold is used in another successful process for making cast iron pipe.[30] Slow cooling in the mold prevents surface chill in these castings.

Additional applications of centrifugal cast irons include cylinder liners, valve seat inserts, and other cylindrical castings.

Effects of Alloying Elements. — The elements nickel, chromium, and molybdenum are the most widely used in alloy cast irons. Vanadium is used to a lesser extent and additions of copper and aluminum are coming into favor for certain special compositions. The qualitative effects of these elements have been summarized in Table LX. It is important to note that this information applies only to gray irons with moderate additions of alloying elements, and not to the special high alloy irons.

It is common practice to add these elements, in suitable form, at the spout or runner during tapping or in the ladle. They may, however, be present in special pig irons or in other portions of the charge.

The compositions of several typical high strength alloy irons are listed in Table LXI.[24] Tensile strengths in the range 40–80,000 lb. per sq. in. are possible using alloy additions compared with 40–50,000 lb. per sq. in. for non-alloyed high strength cast irons. One of the advantages of alloy over non-alloy high strength irons is the possibility of using somewhat higher total carbon contents (hence higher graphite contents) without sacrificing strength when proper alloy additions are made. Considering irons of the same strength, those with higher graphite contents are likely to have better machinability and frictional characteristics.

In discussing alloying elements in cast iron, a characteristic which may be called section sensitivity is a useful criterion.[11] A composition which is relatively free from section sensitivity is one which has a minimum range of hardness and other properties when cast in sections of variable size. The possibility of applying a single heat

TABLE LXI.   TYPICAL COMPOSITIONS OF ALLOY CAST IRONS

Application	Total C	Si	Mn	Cr	Cu	Mo	Ni
Automotive-cylinder blocks	3.35	2.20	0.75	0.35	—	—	0.70
	3.25	2.25	0.60	0.30	—	—	0.25
	3.25	2.00	0.70	—	0.60	—	—
	3.20	2.15	0.70	0.50	—	0.50	0.50
	3.25	2.20	0.60	0.50	1.00	1.00	0.20
	3.35	2.00	0.70	0.35	—	—	1.75
Brake drums	3.30	2.25	0.65	0.30	—	—	0.25
	3.25	2.00	0.60	—	1.00	0.50	—
	3.40	2.40	0.70	—	—	0.75	1.00
	3.20	1.90	0.70	0.50	—	—	1.50
Crankshafts	2.50	2.50	1.00	—	—	1.00	1.00
	2.80	2.10	0.70	—	—	0.60	1.25
	3.00	2.00	0.80	—	—	0.75	1.50
	1.50	0.95	0.70	0.45	1.75	—	—
Sheet-metal forming dies ..	3.25	1.50	1.25	—	—	0.70	—
	3.50	1.50	0.50	—	V0.20	—	1.75
	3.00	2.20	0.75	0.35	—	0.60	2.00
	3.00	1.45	0.90	0.90	—	—	3.00
Machine-tool beds	3.30	1.60	0.60	0.25	—	0.20	1.00
	3.00	1.00	0.75	0.35	—	—	1.25
	2.90	1.90	0.90	—	—	—	1.50
Hot-strip mill rolls	3.00	0.60	0.25	—	—	0.35	—
	3.00	0.55	0.20	0.25	—	0.25	2.25
	3.10	0.60	0.20	0.40	—	0.40	3.50
	3.50	0.90	1.30	2.00	—	—	4.50
	3.40	0.60	0.25	1.30	—	0.40	4.50
Grate bars	3.25	2.00	0.70	1.00	—	—	—
	3.60	1.70	0.50	—	V0.15	—	—
	3.30	1.50	0.40	0.60	—	—	1.50

Courtesy of Paul D. Merica, Trans. Am. Inst. Mining and Met. Eng.[24]

to castings of variable size, or obtaining good machinability and other desirable properties in all sections of a given casting are obvious advantages in using an iron having low section sensitivity.

An effective method for determining this property is the use of a test section of variable thickness, such as the step-bar test illustrated

in Fig. 212. The black areas in these diagrams indicate chilled metal of high hardness and poor machinability.

One of the reasons why nickel is the foremost alloying element used in gray irons is its capacity for reducing section sensitivity. Comparing the first and last bars of Fig. 212, each having a mini-
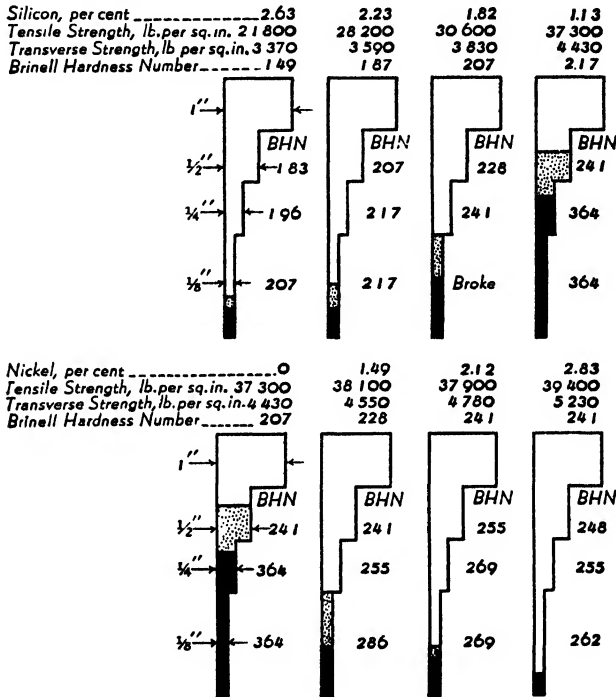
FIG. 212. — STEP-BAR TEST COMPARING THE EFFECTS OF SILICON AND NICKEL IN REDUCING CHILL IN THIN SECTIONS OF IRON CASTINGS. (*Courtesy of P. D. Merica, Proc. Am. Soc. for Testing Mats.*[23])

mum tendency to chill in light sections, the general hardness, strength, and uniformity of the low silicon iron containing nickel is superior to the high silicon iron. From these and other published data it is clear that nickel promotes graphitization without causing excessive softness and weakness. This is due, in part, to the fact that nickel tends to maintain the carbon content of the matrix near the eutectoid composition; furthermore, microscopic evidence indicates that the form and distribution of the graphite is more favorable in nickel and certain other alloy irons than in the high silicon irons.

Under ordinary conditions the graphitizing effects of nickel and silicon are additive, with nickel being about one-half as powerful as a corresponding amount of silicon considering graphitization at temperatures above the eutectoid transformation. Nickel acts very much more slowly than silicon at temperatures below the eutectoid transformation.

It is well known that Brinell hardness depends on the nature of the matrix of cast irons to a greater extent than does the tensile strength. Nickel and other elements such as molybdenum and chromium strengthen and harden the constituents normally present in the matrix by reducing the particle size of the pearlite and forming solid solutions with the ferrite as already described for alloy steels.

Chromium additions are often made to balance the graphitizing effects of high silicon or nickel, or both, in order to maintain or improve the hardness and wear resistance. Chromium increases the stability of the carbides to such an extent that additions of a few tenths of one per cent may be sufficient to give a mottled or white structure in light sections which would otherwise be gray. On the other hand, small additions may be used to reduce the tendency towards coarse graphite in heavy sections. In applications such as cams in which hard surfaces are desirable, chromium is a useful addition for controlling the depth of chill. Elements such as chromium, vanadium, molybdenum, and manganese tend to stabilize the carbides in the matrix upon heating to elevated temperatures as required in certain types of service and in heat treatment.

As indicated by the typical compositions of Table LXI, chromium is often used with nickel and molybdenum in alloy gray irons. As in alloy steels, the combination of nickel and chromium in suitable proportions gives very favorable results. In cast irons this combination has the effect of reducing section sensitivity.

In addition to its strong tendency to stabilize the carbides, vanadium is said to be more effective than chromium in reducing the flake size and improving the uniformity of the graphite. As in the case of alloy steels, small additions of vanadium are generally used along with other alloying elements in special high grade compositions.

Molybdenum additions impart properties somewhat similar to those obtained by balanced additions of nickel and chromium. Molybdenum strengthens gray irons by refining and hardening the matrix and improving the characteristics of the graphite. When present in the amounts ordinarily used, its influence on the total amount of graphite present in gray irons is relatively small compared with either nickel or chromium. Molybdenum, like chromium, is a carbide stabilizer and may be used to increase the hardness and toughness of white irons. It has been observed that molybdenum is particularly effective in reducing graphitization during heat treatment or service at elevated temperatures. When added to base metal of suitable composition, molybdenum has been found to reduce section sensitivity, although not as effectively as nickel.

High alloy irons of the heat and corrosion resistant type are not included in this chapter, and it will be well to remember that the

FIG. 213. — HEAT TREATED ALLOY WHITE CAST IRON. (Etched with picral.)   X1500.

present discussion applies to relatively small additions of alloys to gray and white irons. In many cases large additions of alloys give results which are completely reversed; for example, very high silicon irons are hard and brittle rather than coarsely graphitic.

An intermediate alloy class of hard irons is represented by a nickel-chromium iron of the following composition: 4.0–6.0% Ni, 1.25–2.5% Cr, 0.8–1.15% Si, 2.75–3.5% C. This alloy, known as Ni-hard, has greater hardness, abrasiveness, and toughness than ordinary white or chilled irons.[31] Another type of iron whose resistance to abrasion is outstanding contains approximately 6 to 10% molybdenum as the principal hardening element. Irons in this general classification may be compared with the air hardening

steels since in both cases alloying elements are present in sufficient amounts to produce a martensitic matrix upon cooling in air from above the critical temperature range. In the cast irons the high carbon content is responsible for the presence of massive cementite associated with the martensite, resulting in a very high hardness and resistance to abrasion. The structure of a high molybdenum type white iron containing martensite needles in the matrix is shown in Fig. 213. Its hardness is nearly 1000 Vickers-Brinell.

The use of copper in cast irons in amounts up to 3% is increasing. Several copper alloy irons are used in Ford cylinder blocks, fly-wheels, push rods, and camshafts. In a general way the effect of copper on cast iron is similar to that of nickel.[32] The use of copper in certain special compositions is described at the end of this chapter.

Aluminum is an important addition to special cast irons to be surface hardened by nitriding.[33, 34] Although it has a graphitizing effect similar to that of nickel, its principal function is to increase the hardness of the nitrided surface, as in ordinary nitriding steels. A typical composition for centrifugally cast cylinder liners is: 2.60% C, 1.35% Cr, 1.00% Al, 0.20% Mo, 2.50% Si, 0.50% Mn.[33]

No attempt has been made to cover the entire range of composi-tions of alloy irons available. Because of the large number of indi-vidual producers and the comparatively short time during which alloy irons have been made, there has been less tendency towards standardization than in the field of alloy steels. Data on a wide vari-ety of compositions are included in " Alloy Cast Irons " published by the American Foundrymen's Association. Phillips [36] also gives compositions and properties of cupola and electric furnace alloy irons.

As in the case of alloy versus plain carbon steels, the use of alloy-ing elements is confined to a small percentage of the total output; nevertheless, the importance of their applications justifies the atten-tion which alloy irons receive. Merica [23] has estimated that be-tween 10 and 15% of the gray and chilled iron castings produced in 1936 were alloyed, and that about 15% of the production of all iron (excluding malleable) is of improved or high-duty quality, about one-half of which is consumed in the automotive industry.

Heat Treatment of Cast Iron. — Many of the important applications of cast iron are rather complicated castings such as automobile cylinder blocks, lathe beds, and frame structures for various machines. It is important that the machined surfaces of such castings remain flat and in alignment despite the fact that relief of casting stresses during machining and thereafter tends to produce distortion. To overcome this difficulty the raw or partially machined castings may be stored for several months to relieve the cooling stresses by a natural aging process. A much faster method is to apply a stress relieving annealing treatment, usually at a temperature between 800° and 1000° F. It is desirable to rough machine the castings before annealing and to cool slowly in the furnace.[18, 37, 38]

Annealing for relief of stresses has little effect on the hardness and strength. Heating to higher temperatures under the critical range, 1200° to 1350° F., will produce appreciable softening and improve the machinability of certain irons. Full annealing at temperatures over the critical range, 1400° to 1550° F., produces marked softening and reduction in strength due to graphitization of part or all of the combined carbon. Hard white irons are often annealed at even higher temperatures as described later in this chapter.

In addition to these annealing treatments there is considerable interest in quenching and drawing cast iron for increased hardness, strength, and wear resistance. Since the matrix of a normal cast iron contains considerable silicon and manganese and from 0.50 to 0.80% combined carbon, its hardenability should be good. However, the increase in strength upon heat treatment will not be commensurate with the hardness if the iron contains coarse flake graphite. Excellent results are possible upon hardening irons having fine graphitic structures in the cast condition. As in the heat treatment of steels, alloying elements which improve hardenability are necessary for deep hardening of heavy sections. Most of these elements also tend to prevent graphitization at the heat treating temperature.

The quenching temperature depends somewhat upon the transformation temperature of the matrix, which will be controlled by its composition as in the case of steels. Most irons to be heat treated

have pearlitic or nearly pearlitic matrices which are completely transformed upon heating to 1350° F., and it is seldom necessary to heat gray irons above 1550° F. for quenching. Higher temperatures cause distortion and reduce hardness because of graphitization. The quenching medium is usually oil, but will depend upon the shape and section size of the casting as well as the composition and required hardness. It has been found that the strength and toughness are low after quenching, possibly because of internal stresses, and that a suitable drawing treatment is necessary to develop the high strength desired. Drawing or tempering temperatures up to 1050° F. have been found effective.[18, 39, 40]

Heat treated cast iron automotive camshafts have been produced in considerable numbers using the oxyacetylene flame hardening process for hardening the cam surfaces.

Heat treated gray iron cylinder liners have also met with success, particularly from the standpoint of wear resistance. The following composition has been used for liners for truck and motorcoach engines: 3.1–3.4% T.C., 0.75–0.90% C.C., 1.9–2.1% Si, 0.55–0.75% Mn, under 0.20% P, under 0.10% S, 1.8–2.2% Ni, 0.55–0.75% Cr. The improvement in the hardness and strength of this iron upon heat treatment is illustrated in Table LXII by Eddy.[41]

TABLE LXII.   HEAT TREATMENT OF ALLOY CAST IRON
CYLINDER LINERS

Treatment	Tensile Strength lb./sq. in.	Brinell Hardness
As cast	37,000–41,000	——
Aged at 1025°–1040° F. for 2 hrs.	38,000–43,000	217–248
Quenched in oil from 1550° F.	28,000–36,000	570–650
Quenched — drawn at 300° F. for 2 hrs.	33,000–39,000	530–610
” — ” ” 350° F. ” 2 ”	36,000–40,000	500–580
” — ” ” 450° F. ” 1 ”	39,000–43,000	470–530
” — ” ” 600° F. ” 1 ”	45,000–56,000	450–500

Welding of Cast Iron. — Cast iron is inherently difficult to weld because of its sensitivity to cracking and its tendency to form hard spots upon rapid cooling from the welding temperature. As in the case of the higher carbon and alloy steels, careful preheating of the

work and slow cooling after welding are necessary precautions when either cast iron or mild steel welding rods are used. In the case of irons, however, there is the attendant danger of excessive graphitization. Bronze welds are applied at much lower temperatures and without preheating, consequently they are extensively used for repairing iron castings.[37, 42]

Malleable Cast Iron. — As indicated on page 413 the number of malleable iron foundries is small compared with the number of gray iron foundries; nevertheless, the malleable iron foundries in this country have an annual capacity of over 1,000,000 tons and they supply annealed castings to the railway, farm implement, automotive, electric power, hardware, plumbing, and other industries.

Compared with ordinary gray cast irons, malleable iron is softer, easier to machine, stronger, tougher, and much more ductile; hence it occupies a strong position in the field of iron castings in spite of its higher cost.

Malleable iron is made from white iron castings by a high temperature, long time annealing treatment. The original white iron castings are of a low carbon low silicon type which does not readily graphitize when cast in sand molds. It is difficult to melt irons of this composition in a cupola because higher temperatures are required than for gray irons, therefore air furnaces are used. The cost of melting white irons is greater than for cupola melted gray irons, and the white iron is more difficult to cast into intricate shapes.

Principle of Malleableization. — White iron castings contain the microconstituents indicated in the ordinary Fe-Fe_3C equilibrium diagram for alloys containing 2.0 to 3.0% carbon; namely, cementite and pearlite. Because of the comparatively low carbon content of white irons to be malleableized, the amount of massive cementite will usually be less than in chilled sections of gray irons. A typical example is shown in Fig. 214.

The tendency for decomposition of the cementite by means of the reaction,

$$Fe_3C \rightarrow 3Fe + C \text{ (graphite)},$$
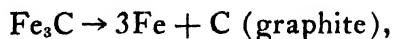
is not great enough to cause graphitization upon solidification of these irons; however, reheating to a red heat for a sufficiently long

time, followed by slow cooling, will permit the reaction to go to completion giving the vastly different structure of Fig. 215, a fully annealed malleable iron. The graphite present in this structure is different from any found in the gray iron photomicrographs at the beginning of this chapter. The nodular graphite formed by dissociation of cementite by annealing is called temper carbon. The

Left. FIG. 214. — SAND CAST WHITE IRON FOR MALLEABLEIZING CONTAINING 2.55% C, 0.80% SI. (Etched with picral.) Brinell 429. X100.
Right. FIG. 215. — MALLEABLE CAST IRON. (Etched with nital.) Brinell 109. X100.
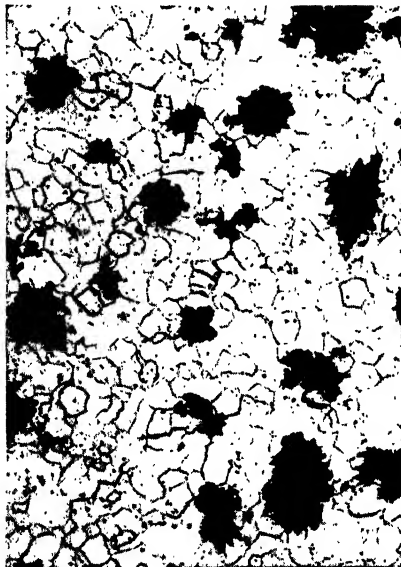
matrix in which it is held is ferrite containing silicon, manganese, and phosphorus in solid solution. Temper carbon gives malleable iron a dark fracture from which the term black heart malleable iron is derived. White heart malleable iron is a decarburized form to be described later.

Heat Treatment for Producing Black Heart Malleable Iron. — The following typical malleableizing cycle, in five operations, will be used for further study of the process.

(1) Heating to 1600° F. in 36 hours.

(2) Holding at 1600° F. for 48 hours. When this stage is com-

plete all of the massive cementite present will have dissoci-
ated to temper carbon held in a matrix of austenitic iron
containing carbon in solid solution. The carbon saturated
austenite is stable at this temperature, therefore a longer
soaking time cannot precipitate more graphite.

(3) Cooling slowly to 1300° F. at the rate of 10° F. per hour.
As the temperature drops, the solubility of carbon in the
austenitic matrix diminishes along Acm until it reaches the
eutectoid composition at about 1350° F. The Fe_3C
which is precipitated from the austenite is graphitized dur-
ing cooling. Austenite of eutectoid composition trans-
forms to pearlite in the usual way at Ar_1.

(4) Holding at 1300° F. for 24 hours. Although the graphitiz-
ing reaction is much slower at this temperature than in
operations 2 and 3, the cementitic portion of the pearlite
will graphitize during this period, giving a final structure
similar to Fig. 215.

(5) Cooling to room temperature. No further change takes
place.

The above process for malleableization requires about 7 days as
carried out in large batch type furnaces. The castings are packed
in iron pots with sand, cinders, or other inert materials. A charge
of this type, weighing as much as 100 tons, has a high heat capacity
which makes rapid heating or cooling impossible.

Although this long process is still used for annealing malleable
castings, much progress has been made towards shorter annealing
cycles. Faster heating and cooling are accomplished by using
smaller batch type furnaces, by using continuous car type furnaces
or kilns, by loading the castings on racks without packing materials,
and by using more efficient furnaces in place of the older installations
which are often of the hand fired, coal burning type.[43] Recent fig-
ures show that about 13.5% of the castings are annealed in hand
fired furnaces and that pulverized coal furnaces are the most widely
used.[44]

Faster furnace operation is permissible and economically desir-
able in operations (1), (3), and (5) of the above outline. In op-
eration (3) rapid cooling causes separation of cementite faster than

it is graphitized; however, it has been found that all of the cementite can be decomposed in operation (4) with a net saving of time.

The time required for operation (2) can be considerably reduced by using higher temperatures between 1700° and 1850° F. In a typical short cycle anneal carried out in an electric furnace the cast-, ings are heated to 1750° F. in 4 hours and held for only 10 hours before cooling. The maximum temperature which may be used is limited by warping and scaling of the castings. Various short cycles have been worked out, and total annealing times of 28 to 48 hours are quite common.[45, 46]

It is apparent that higher silicon and carbon contents would promote graphitization and speed up the malleableizing process; however, the chemical composition is controlled by the necessity for preventing the formation of flake graphite in the raw castings.

White Heart Malleable Iron. — The atmosphere in the annealing furnace must be only slightly oxidizing to produce black heart malleable iron. Usually there will be a skin of decarburized metal, free from graphite, at the surface of the casting. On the other hand, white heart malleable iron is made by packing in mill scale or other oxidizing materials in order to oxidize all the graphite, leaving a ferritic microstructure. This process is seldom used in this country because the ductility of the product is much inferior to black heart malleable iron.

Composition and Properties of Malleable Iron. — The compositions of white irons for malleableizing have been classified by Schwartz [47] in the following manner:

% C	% Si	Application
2.0–2.2	1.0	For highest strength and ductility
2.3–2.6	0.8–0.9	For a fair combination of all properties
2.8–3.0	0.7	Where prevention of foundry difficulties is a major factor

(The carbon contents are reduced to 1.2–2.0% during annealing.)

The silicon contents may vary from the above depending upon the section size. Heavy castings, for example, may require lower silicon contents to prevent the formation of flake graphite in the raw casting. On the other hand, small low carbon castings may contain 1.6% silicon to facilitate graphitization upon annealing.

The manganese content is generally under 0.3% in air furnace and 0.6% in cupola white irons. The sulphur content is under 0.15% in air furnace and 0.25% in cupola irons. Phosphorus may be present up to 0.2%.[10]

TABLE LXIII.   MECHANICAL PROPERTIES OF MALLEABLE IRON

Air Furnace Malleable	Tensile Strength lb./sq. in.	Yield Strength lb./sq. in.	Elongation in 2 in., %
Results of 5,000 tests [48]	{50 000 {59 000	{34 000 {40 000	{10 {30
Average results of 20,000 tests [10]	54 000	36 000	18
A.S.T.M. Spec. A47–33 — Grade 32510	50 000	32 500	10
A.S.T.M. Spec. A47–33 — Grade 35018	53 000	35 000	18
Cupola Malleable			
Cupola malleable iron for fittings [49]	48 000	34 000	9
Manufacturers Standardization Society of the Valve and Fitting Industry — SP22 — 1931	40 000	30 000	5

Considering that the modulus of elasticity of malleable iron is about 25,000,000 lb. per sq. in. and that the yield strength is over 30,000 lb. per sq. in., it is apparent that its strength characteristics for design purposes are more nearly comparable to those of steel castings than to those of gray cast irons. Moreover, malleable iron is much superior to steel castings in machinability; in fact, malleable iron is considered to be the easiest to machine of all commercial ferrous alloys. Although temper carbon contributes greatly to the machinability of malleable iron, it is much less detrimental to strength, elasticity, and ductility than is the flake graphite of gray cast iron.

Malleable iron is unusual in that its ductility and strength increase and decrease simultaneously, rather than inversely as in most other alloys.

The hardness of malleable iron ranges from 100 to 145 Brinell. Its impact toughness is about 8 foot pounds in the Charpy test, which is much higher than for gray irons. These properties explain its successful application as structural castings requiring fair strength and shock resistance, while gray irons and pearlitic malleable irons

are more suitable for products in which wear resistance is the most important consideration.

About 8% of the production of white irons for making malleable iron is melted in cupolas. Cupola malleable iron is in the higher carbon range which is characterized by greater fluidity and lower
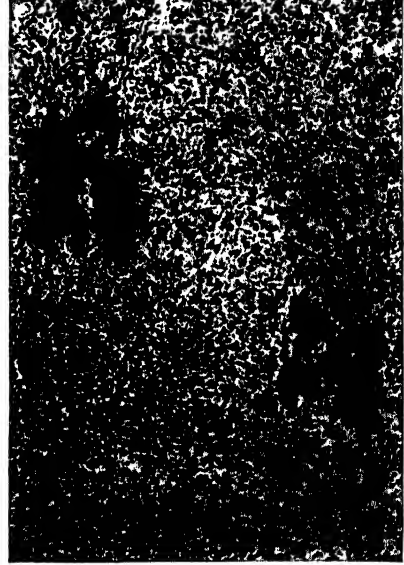
Left. FIG. 216. — PARTIALLY MALLEABLEIZED WHITE IRON. (Etched with picral.) Brinell 183. X100.

Right. FIG. 217. — PEARLITIC MALLEABLE CAST IRON. (PROMAL.) (Etched with picral.) Brinell 180. X500.

shrinkage. Although cupola malleable iron is less ductile, it is at least equal to air furnace malleable iron in pressure tightness and is less likely to become brittle after galvanizing.[10] These characteristics are responsible for the extensive use of cupola malleable iron for pipe fittings.

Malleable iron is difficult to weld. Low temperature bronze welding is less likely to cause embrittlement than gas or arc welding.[50]

Pearlitic Malleable Iron. — Although there are but a few grades of ferritic malleable irons, numerous pearlitic types and alloy grades have been developed to supply the demand for higher strength alloys

having the general characteristics of malleable iron.   In the " Symposium on Pearlitic Malleable Cast Iron " [51] the commercial types are classified by processes and by metallographic characteristics. The following symbols were required to include all the variations: Groups 1 to 6, Class A and B, Divisions I and II, alpha and beta Families, and Types a to d.   Only three types will be included in the present discussion.

The pearlitic malleable irons are produced by raising the combined carbon content of the matrix from practically zero to 0.30–0.90%.   This may be accomplished in a number of different ways by variation of the malleableizing annealing treatment.   Probably the simplest method for strengthening the matrix is to arrest the graphitizing reaction during the final stages (operation 4 on page 455).   A typical structure obtained by this method is shown in Fig. 216.   The white areas around the black temper carbon nodules are ferrite, and the remainder of the dark material is pearlite.   The name " bull's eye " malleable iron has appropriately been applied to this structure.   Its increased hardness and strength compared with ferritic malleable iron are obtained at a sacrifice of ductility.

A structure obtained by the Promal process is illustrated in Fig. 217.   Because of the high magnification used to resolve the sorbitic structure of the matrix, only two areas of temper carbon are included in the field.   Irons of this type are produced from fully annealed malleable by a further heat treatment involving re-solution of carbon from the temper carbon into the matrix at temperatures above Ac_1, followed by quenching in oil and drawing to a sorbitic condition.   The mechanical properties ordinarily obtained are 50,-000 lb. per sq. in. minimum yield point, 70,000 lb. per sq. in. minimum tensile strength, 10 to 14% elongation in 2 inches, and 170 to 190 Brinell hardness.   The strength may be increased to 90,000 lb. per sq. in. by varying the heat treatment, or to 100,000 lb. per sq. in. by the use of alloys.[51]   Among the principal applications of this alloy are chains and chain products for power transmission and for conveyance of materials.   Its increased hardness provides a marked improvement in the wear resistance over that of ordinary malleable iron.

Z-Metal is representative of malleable irons which retain com-

bined carbon in the matrix through the use of alloying elements, such as manganese, which retard graphitization in the final stages of the annealing cycle. The microstructures obtained are similar to Fig. 217. The following range of mechanical properties is available through variation in composition and treatment.[51]

Tensile Strength, lb./sq. in.	Yield Point, lb./sq. in.	Elongation in 2 in., %	Brinell Hardness
70 000	50 000	16–18	179
80 000	56 000	12–14	191
90 000	62 000	9–11	207
100 000	70 000	6–8	217

Further details of these interesting and increasingly important pearlitic malleable irons will be found in references 43, 51, and 52.

Graphitic Steel Castings. — The tendency towards lower carbon contents and higher strengths in malleable cast irons has reached the stage where certain products are difficult to distinguish from high carbon cast steels. The neutral zone which formerly existed between steels and irons, 1.4 to 2.0% carbon, is now occupied by commercial alloys which are structurally similar to pearlitic malleable irons. Although they contain temper carbon, these alloys have become known as alloy cast steels.[53, 54, 55, 56] They are used for certain important parts of Ford automobiles, including the crankshaft and brake drums.

The application of a cast metal to an automotive crankshaft was considered out of the question ten years ago; nevertheless, the graphite containing alloy castings developed for the purpose are now highly successful. The data in Table LXIV indicate that the utility of the cast metal is not reflected by its static tensile properties. Results of impact tests are equally misleading. The reliability of the castings in service has been attributed to their good endurance characteristics, which include reduced " notch sensitivity "[57] and high damping capacity, compared with wrought steels used for the same purpose. Good wear resistance and low cost are further advantages of the cast alloys.

These castings are essentially white as cast and must be heat treated before using. The treatment for the brake drum, for ex-

TABLE LXIV. COMPOSITIONS AND PROPERTIES OF GRAPHITIC
ALLOY CAST STEELS

Application	C	Si	Mn	Cr	Cu
Crankshaft [53, 56] ...	1.35–1.60	0.85–1.10	0.6–0.8	0.4–0.5	1.5–2.0
Pistons [56]	1.30–1.70	0.9–1.3	0.6–0.8	0.08–0.15	2.5–3.0
Brake drums [55]	1.55–1.70	0.9–1.1	0.7–0.9	——	2.0–2.25
Fig. 218	1.69	1.29	0.62	——	1.81

	Copper Alloy Cast Steel for Crankshafts [53]	S.A.E. 1045
Tensile strength, lb./sq. in.	107,500	116,000
Yield strength, lb./sq. in.	92,000	84,000
Elongation in 2 in., %	2.5–3.0	16
Reduction in area, %	2.0–2.5	46
Brinell hardness	269	248

ample, consists of heating to 1650° F., holding 30 minutes at temperature, cooling rapidly to 1400° F., cooling at the rate of 90° per hour to 1200° F., and cooling to 1000° F. in one hour. Typical mechanical properties after heat treatment are: elastic limit 70,000 lb. per sq. in., tensile strength 85,000 lb. per sq. in., elongation in 2 inches 7%, and Brinell hardness 220.[55]

A somewhat similar heat treatment, using lower cooling rates, was applied to the iron whose structure is shown in Fig. 218. The manganese present is said to be responsible for the retention of sufficient combined carbon to give the pearlitic matrix. The low total carbon content is responsible for the relatively small amount of graphite or temper carbon present.

Wrought Graphitic Steel. — Hot working steels should not properly be included in this chapter. Nevertheless, the newly developed materials known as graphitic steels are similar to the casting alloys just described in their metallography and chemical compositions and are introduced at this point rather than in Chapter VII.

By carefully adjusting the silicon and carbon content it is possible to produce a steel which will not graphitize upon casting in ingot molds and rolling or forging to shape, but will form temper carbon upon subsequent heat treatment. The structure and properties of
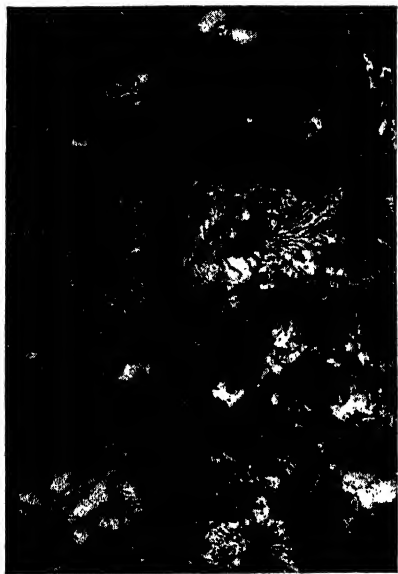
FIG. 218. — GRAPHITIC ALLOY CAST STEEL. (SEE TABLE LXIV.) (Etched with picral.) Brinell 197. X100.

the matrix may be varied within wide limits depending upon the final treatment. In some cases the microstructures are very similar to Figs. 216, 217, or 218, while in the fully hardened condition the matrix is martensitic. The following is a typical composition of an alloy of this type known as " Graph-sil ": 1.51% C, 0.96% Si, 0.40% Mn, 0.015% P, 0.014% S.[58]

The graphite present after annealing contributes greatly to the machinability, and after hardening it aids lubrication at bearing surfaces. Some of the applications which have already been found for this type of material include bushings, gages, brake drums, piston rings, and dies. Remarkable wear resistance and service life have been reported in certain applications.[58]

Hot Rolling of Cast Iron. — Hot rolling of graphitic alloys is not restricted to the above class of materials. It has been found that ordinary white cast irons may be hot worked under certain conditions; thus the many desirable properties ordinarily associated with castings of malleable iron are becoming available in the form of sheets and other rolled shapes.[59]

REFERENCES

1. " Roll Call Shows Slight Change," The Foundry, **63**, 48 (June 1935)
2. Pinel, M. L., Read, T. T., and Wright, T. A., " Composition and Microstructure of Ancient Iron Castings," Trans. Am. Inst. Mining and Met. Eng., **131**, 174 (1938)
3. Stoughton, B., " Metallurgy of Iron and Steel," McGraw-Hill Book Company, New York (1934)
4. Hurst, J. E., " Melting Iron in the Cupola," The Penton Publishing Co., Cleveland (1929)

5. Mahin, W. E., and Hamilton, J. W., "A Proposed Standard Classification of Graphite in Gray Cast Iron," Proc. Am. Soc. for Testing Mats., **37**, Part II, 52 (1937)

6. Bolton, J. W., "Gray Cast Iron," The Penton Publishing Co., Cleveland (1937) — and — "Gray Cast Iron," The Foundry, A series of 23 articles from **63**, April 1935, to **65**, Feb. 1937

7. Bolton, J. W., "Graphitization and Inclusions in Gray Iron," Trans. Am. Foundrymen's Assoc., **45**, 467 (1937)

8. Greiner, E. S., Marsh, J. S., and Stoughton, B., "Alloys of Iron and Silicon," The Engineering Foundation, McGraw-Hill Book Company, New York (1933)

9. Boyles, A., "The Freezing of Cast Iron," Trans. Am. Inst. Mining and Met. Eng., **125**, 141 (1937)

10. Sisco, F. T., "Alloys of Iron and Carbon, Vol. II — Properties," McGraw-Hill Book Company, New York (1937)

11. Piwowarky, "Certainty of Results as the Basis for the Manufacture of High Test Gray Iron," Trans. Am. Foundrymen's Assoc., **42**, 705 (1934)

12. DiGiulo, A., and White, A. E., "Factors Affecting the Structure and Properties of Gray Cast Iron," Trans. Am. Foundrymen's Assoc., **43**, 531 (1935)

13. Parke, R. M., Crosby, V. A., and Herzig, A. J., "Graphitization of Gray Cast Iron," Metals and Alloys, **9**, 9 (1938)

14. Spencer, W. H., "Chilling Properties of Cast Iron," Trans. Am. Foundrymen's Assoc., **42**, 508 (1934)

15. Murphy, D. W., Wood, W. P., and Girardi, D., "The Rate of Austenite Transformation in Cast Iron," Trans. Am. Soc. for Metals, **26**, 173 (1938)

16. Cone, E. F., "The Brackelsberg Furnace as a Foundry Melting Unit," Metals and Alloys, **9**, 105 (1938)

17. Culling, C. R., "The Experience of a Gray Iron Jobbing Foundry with the Rocking-Type Electric Furnace," Trans. Am. Foundrymen's Assoc., **41**, 519 (1933)

18. Walther, H. H., "Electric Furnace Cast Iron Practice," Trans. Am. Foundrymen's Assoc., **42**, 32 (1934)

19. "Reports of Gray Iron Committees on Recommended Practices — Analyses for Classes of Castings," Trans. Am. Foundrymen's Assoc., **39**, 59 (1932)

20. "Report of Subcommittee XV, Committee A–3, on Impact Testing," Proc. Am. Soc. for Testing Mats., **33**, Part I, 87 (1933)

21. "Standard Specifications for Gray-Iron Castings — A.S.T.M. Designation A48–36," Am. Soc. for Testing Mats., Standards, Part I, 457 (1936)

22. Campbell, H. L., "Relation of Properties of Cast Iron to Thickness of Castings," Proc. Am. Soc. for Testing Mats., **37**, Part II, 66 (1937)

23. "Symposium on Cast Iron," Proc. Am. Soc. for Testing Mats., **33**, Part II, 115 (1933)

24. Merica, P. D., "Progress in Improvement of Cast Iron and Use of Alloys in Iron," Trans. Am. Inst. Mining and Met. Eng., **125**, 13 (1937)

25. Coyle, F. B., "Properties and Heat Treatment of Cast Iron for Diesel Engines," Trans. Am. Soc. for Steel Treating, **12**, 446 (1927)

26. MacPherran, R. S., "Cupola High Test Cast Iron," Trans. Am. Foundrymen's Assoc., **45**, 343 (1937)

27. Sherwin, L. M., and Kiley, T. F., "Cupola High-Test and Alloy Irons in a Machine Tool and Gray Iron Jobbing Foundry," Trans. Am. Foundrymen's Assoc., **39**, 115 (1931)

28. Lemoine, R. P., "Cupola High Test Cast Iron," Trans. Am. Foundrymen's Assoc., **42**, 745 (1934)

29. Walls, F. J., "Permanent Mold Castings," Trans. Am. Foundrymen's Assoc., **42**, 427 (1934)

30. "Centrifugal Casting Symposium," Trans. Am. Foundrymen's Assoc., **43**, 443 (1935)

31. Vanick, J. S., "An Improved Nickel-Chromium Hardened Chilled Cast Iron," Trans. Am. Inst. Mining and Met. Eng., **105**, 53 (1933)

32. Eastwood, L. W., Bousu, A. E., and Eddy, C. T., "Copper and Copper Manganese Gray Cast Iron," Trans. Am. Foundrymen's Assoc., **44**, 51 (1936)

33. Ross, E. F., "Nitriding Iron Cast Centrifugally," The Foundry, **63**, 26 (March 1935)

34. Homerberg, V. O., and Edlund, D. L., "Nitricastiron," Metals and Alloys, **5**, 141 (1934)

35. "Alloy Cast Iron," Am. Foundrymen's Assoc., Chicago (1939)

36. Phillips, G. P., "Impact Resistance and Other Physical Properties of Alloy Gray Cast Irons," Trans. Am. Foundrymen's Assoc., **43**, 125 (1935)

37. Bornstein, H., and Bolton, J. W., "Cast Iron," Metals Handbook, 617, Am. Soc. for Metals, Cleveland (1939)

38. McElwee, R. G., "Heat Treatment of Cast Iron," Trans. Am. Foundrymen's Assoc., **43**, 27 (1935)

39. Hurst, J. E., "Heat Treatment of Cast Iron by Hardening and Tempering," Trans. Am. Foundrymen's Assoc., **44**, 373 (1936)

40. Ballay, M., and Chavy, R., "Some Applications of Heat Treated Cast Irons in France," Trans. Am. Foundrymen's Assoc., **44**, 161 (1936)

41. Eddy, W. P. Jr., "Heat Treated Gray Iron Cylinder Liners," Trans. Am. Foundrymen's Assoc., **42**, 129 (1934)

42. Spraragen, W., and Claussen, G. E., "Welding Cast Iron — A Review of the Literature to November 1, 1936," Supplement to The Welding Journal, **16**, 2 (March 1937)

43. Schwartz, H. A., "The Field of Malleable Cast Iron," Trans. Am. Foundrymen's Assoc., **8**, 23 (1937)

44. Bean, W. R., "Modern Malleable Practice," Metal Progress, **33**, 286 (1938)

45. White, A. E., and Schneidewind, R., "Malleable Cast Iron Annealed in a Short Cycle," Metal Progress, **25**, 25 (June 1934)

46. Bryce, J. T., and Schwab, H. G., "Faster Annealing of Malleable Cast Iron," Metal Progress, **33**, 35 (1938)

47. Schwartz, H. A., "Malleable Iron — Recent Progress," Metals and Alloys, **2**, 143 (1931)

48. "Symposium on Malleable Iron Castings," Proc. Am. Soc. for Testing Mats., **31**, Part II, 317 (1931)

49. Riggan, F. B., "Notes on the Production of Cupola Malleable Castings," Trans. Am. Foundrymen's Assoc., **43**, 427 (1935)

50. Spraragen, W., and Claussen, G. E., "Welding Malleable Cast Iron — A Review of the Literature to December 1, 1936," Supplement to The Welding Journal, **16**, 10 (June 1937)

51. "Symposium on Pearlite Malleable Cast Iron," Am. Soc. for Testing Mats., Philadelphia (1936)

52. Sauveur, A., and Anthony, H. L., " Malleable Castings," Trans. Am. Soc. for Metals, **23,** 409 (1935)
53. Cone, E. F., " The Story of the Ford Cast Crankshaft," Metals and Alloys, **6,** 259 (1935)
54. Cone, E. F., " The Lincoln-Zephyr Cast Alloy Piston," Metals and Alloys, **7,** 85 (1936)
55. Cone, E. F., " The Metallurgical Story of Ford Brake Drums," Metals and Alloys, **8,** 303 (1937)
56. McCarroll, R. H., and McCloud, J. L., " Ford Alloy Castings," Metal Progress, **30,** 33 (Aug. 1936)
57. Russell, H. W., " Resistance to Damage by Overstress," Metals and Alloys, **7,** 321 (1936)
58. Bonte, F. R., and Fleischmann, M., " Developments in Graphitic Steel for Tools and Dies," Metal Progress, **31,** 409 (1937)
59. Forbes, D. P., " New Cast Irons Heat Treated, Rolled," Metal Progress, **33,** 137 (1938)

HIGH ALLOY STEELS

High alloy steels containing 12 per cent or more of alloying elements are useful because of the special properties or combinations of properties which they develop. Their principal applications depend upon resistance to corrosion at normal and moderately elevated temperatures, resistance to oxidation at high temperatures, resistance to wear as tools and dies or machine parts, special electrical properties such as high permeability, or special physical properties such as low coefficient of expansion. The usual mechanical properties such as tensile strength and ductility vary widely with the composition and, in many cases, with mechanical and heat treatment. The desirable mechanical properties of ordinary carbon and alloy steels can, to a large extent, be duplicated by proper selection and treatment of a stainless steel.

One high alloy material, austenitic manganese steel, has already been discussed in Chapter XI. High alloy tool steels are included in Chapter XV. In this chapter the special electrical and nonexpanding alloys will be mentioned briefly, leaving the heat and corrosion resisting compositions as the principal topic.

Iron-Nickel Alloys. — The alloys of iron with high nickel contents are the basis for many of the special compositions used in the electrical industry for communication equipment. Invar and other alloys having low or special coefficients of thermal expansion also belong to the iron-nickel system. Marsh[1] has recently described all iron-nickel alloys in great detail in a treatise of the " Alloys of Iron Monograph Series."

A constitution diagram for iron-nickel alloys, as given by Merica in " Metals Handbook "[2], is shown in Fig. 219. Nickel and iron form solid solutions in all proportions. Nickel, which has the same type of crystal structure as gamma iron, suppresses the transformation of gamma to alpha on cooling, thus it tends to stabilize the face-centered cubic phase at low temperatures. In alloys containing less

than about 30% nickel, the alpha phase is formed on slow cooling to room temperature as shown by the lower shaded area of Fig. 219. Upon heating alloys containing the alpha phase, transformation to gamma occurs only at relatively high temperatures. Because of the apparent sluggishness of the gamma-alpha transformation these alloys have become known as the irreversible iron-nickel alloys.

Aside from the use of nickel in structural and machine steels, few applications have been made of iron-nickel alloys in this range.

The magnetic transformation of the reversible alloys is also suppressed, especially in the compositions containing 20 to 30% nickel which have a limited application as castings and forgings for nonmagnetic parts of electrical machinery.

Alloys containing more than about 35% nickel are face-centered cubic (gamma) solid solutions at all temperatures above room temperature and up to their melting

FIG. 219. — THE IRON-NICKEL CONSTITUTION DIAGRAM. (*Courtesy of P. D. Merica, Metals Handbook.*[2])

points, hence they are not subject to grain refinement or hardening by heat treatment. (They may be recrystallized after cold working.) It is in this range, however, that many useful compositions are found. The 36% nickel alloy (Invar) has a very low coefficient of expansion at normal and moderately elevated temperatures.[3] It is widely used for measuring tapes and other measuring devices, watch parts, and certain machine parts.

Elinvar contains about 32% nickel with the addition of 4 to 5% chromium, 1 to 3% tungsten, and smaller amounts of manganese, silicon, and carbon.[4] In addition to a low coefficient of thermal expansion, this composition maintains a constant modulus of elasticity over a temperature range of 0° to 100° F., a property which is highly desirable in hairsprings for watches and springs for other precision instruments.

Certain other iron-nickel alloys with nickel contents between 36 and 46% have specific coefficients of expansion equivalent to those of platinum and certain types of glass, hence they are useful as lead-in wires for lamps and vacuum tubes.

There are a number of iron-nickel alloys in use in the electrical industry which contain up to 85% nickel, and, in some cases, additions of copper, cobalt, and aluminum, which have valuable electrical characteristics. These range from a 50% nickel alloy (Hipernik[5]), which has a low hysteresis loss and very high magnetic permeability at low inductions, to a complex permanent magnet alloy (Alnico) containing approximately 20% nickel, 5% cobalt, and 12% aluminum.[1] Permanent magnet alloys of this type cannot be worked mechanically and are used as castings hardened by a precipitation heat treatment. Magnets made of the precipitation hardening alloys, some of which contain titanium or copper, have great magnetic strength and permanence.*

Alloys known as Permalloy are probably the most widely known of the high nickel electrical varieties. The grade containing 78.5% nickel has very high permeability and low hysteresis loss at low inductions. It is used in telephone and ocean cable work.

Iron-Silicon Alloys. — High silicon cast irons containing 13 to 15% silicon are among the most resistant of all commercially available alloys to sulphuric and nitric acids and many other corrosive chemicals.[6,7] They are hard and brittle and low in tensile strength, hence they are limited in application to cast equipment.

Stainless Steels. — In 1937 about 140,000 gross tons of stainless steel ingots were made in this country, amounting to about 17% of all steel melted in electric furnaces.

The iron-chromium alloys are the basis of modern stainless steels. The corrosion resisting qualities imparted to iron by high chromium contents were recognized and commercialized in England by Harry Brearley shortly before the World War. At about the same time Benno Strauss and Eduard Maurer developed the first iron-chromium-nickel types of stainless steels in Germany. Several years earlier Elwood Haynes of Kokomo discovered the heat and corro-

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* The general subject of magnet steels is not included in this text. The reader is referred to Bullen's " Steel and Its Heat Treatment," Vol. 2.

sion resistance of certain high chromium compositions. (See Chapter XV.) Although nickel extends the range of corrosion resistance of the basic iron-chromium alloys, it has an even stronger influence on their mechanical properties, greatly improving their plasticity, toughness, and welding characteristics.

Both the iron-chromium and the iron-chromium-nickel alloy types may be modified by additions of molybdenum, silicon, copper, and many other elements to develop special characteristics. Carbon and the minor elements present in ordinary steels are also present in stainless steels.

Although the true nature of the protective action of chromium upon the corrosion of iron has not been fully or conclusively established, the oxide film theory is quite generally accepted. An oxide film forms naturally on the surfaces of ordinary irons and steels upon exposure to the atmosphere or to oxidizing reagents and offers more or less protection against further oxidation. After immersion in strong nitric acid this film becomes highly protective and the iron is said to be passive. Passivity of iron produced in this manner is, however, only temporary, whereas the film which is said to form on high chromium-iron alloys upon exposure to the atmosphere is much more stable and is self-healing under normal conditions. In general, the stainless steels are most effective against oxidizing substances which tend to maintain the protective film; however, they are also resistant to a wide variety of non-oxidizing chemicals and to many organic substances. Among the common acids for which ordinary grades of stainless steels are not recommended are hydrochloric, hydrofluoric, sulphuric, and sulphurous. Special compositions have been developed to withstand sulphuric and sulphurous acids and other materials which the standard compositions do not resist.

Iron-Chromium Alloys. — The stainless steels produced by Brearley contained approximately 11 to 14% chromium and 0.30 to 0.40% carbon. It was discovered that their resistance to attack by acids, as well as their hardness and strength, was greatly influenced by heat treatment. In order to understand this behavior and other characteristics of the iron-chromium alloys it is necessary to study the binary iron-chromium equilibrium diagram and certain sections of the ternary iron-chromium-carbon alloy system.

Considering first the binary or carbon-free alloys, as represented in Fig. 220, it is apparent that at very high temperatures, just after solidification, chromium forms a continuous series of alpha (B.C.C.) solid solutions with iron. (Although the special symbol delta is used in the conventional iron-iron carbide diagram to denote the B.C.C. iron phase existing between 1400° and 1535° C., the symbol alpha has been used to designate all B.C.C. solid solution alloys at all temperatures in Fig. 220.) The gamma iron (F.C.C.) phase which is stable between 908° and 1400° C. disappears when the chromium content reaches about 12%. The resulting "gamma loop" at the left in Fig. 220 is characteristic of several other binary systems in which the added element suppresses the F.C.C. or gamma iron phase and tends to stabilize the alpha phase. The alloy content at the right hand margin of the gamma phase region may also be considered to be the limit of solubility of the added element in gamma iron. This limit is approximately 11.5% for the alloys of iron and chromium, 2.5% for molybdenum, 3% for tungsten, 1% for vanadium, 0.3% for phosphorus, 2% for silicon, and 1% for aluminum.

FIG. 220. — THE IRON-CHROMIUM EQUILIBRIUM DIAGRAM. (*Courtesy of A. B. Kinzel and Walter Crafts, The Alloys of Iron and Chromium, Vol. I.*[8])

High chromium-iron alloys with chromium contents to the right of the gamma loop have the B.C.C. structure of ferrite. (See page 482 for discussion of the sigma phase.) Small amounts of carbon increase the chromium content necessary to close the gamma loop; nevertheless, many of the commercial stainless steels have ferritic structures.

Although the chromium content at which the corrosion resisting
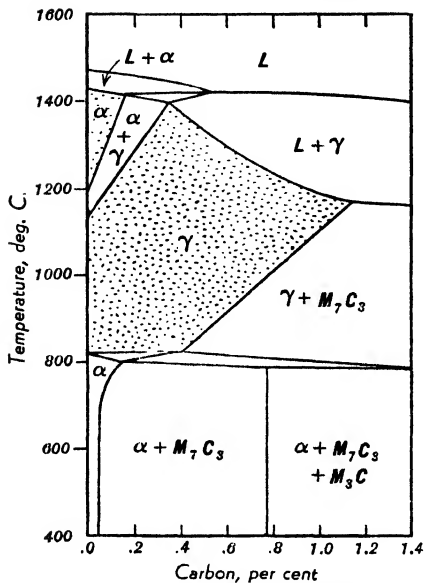
Fig. 221. — Section at 8 % Cr
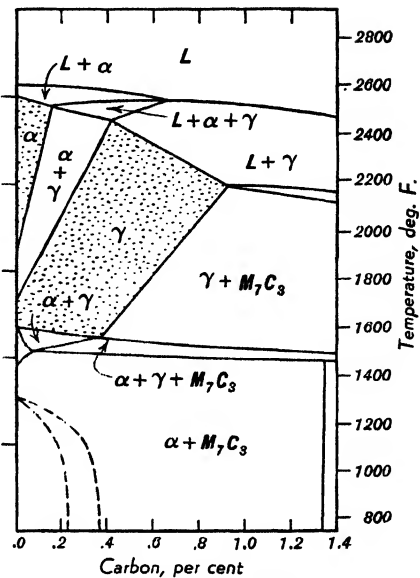
Fig. 222. — Section at 12 % Cr
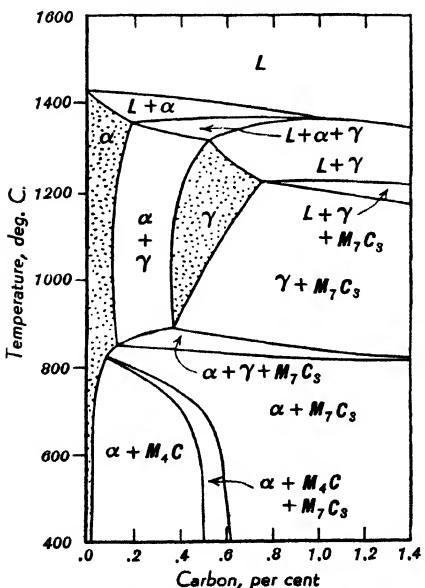
Fig. 223. — Section at 15 % Cr
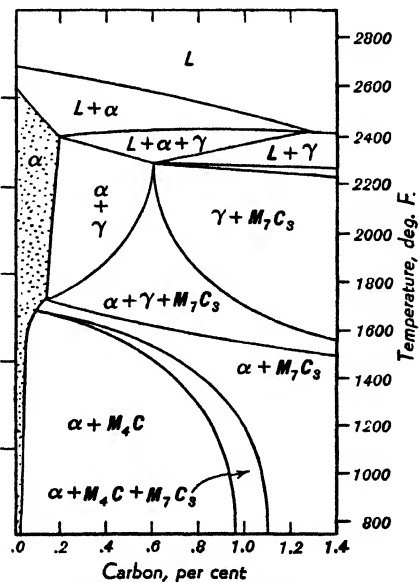
Fig. 224. — Section at 20 % Cr

FIGS. 221 TO 224. — SECTIONS OF THE IRON-CHROMIUM-CARBON EQUILIBRIUM DIA-
GRAM AT CONSTANT CHROMIUM CONTENTS. (*Courtesy of W. Tofaute and Coworkers,
Archiv für das Eisenhüttenwesen.*[9]) (" M " denotes both Cr and Fe in the carbides.)

qualities become fully operative coincides quite well with the closing of the gamma loop at about 12% chromium, the ferritic structure is not in itself the essential factor.

Iron-Chromium-Carbon Alloys. — The binary diagram of Fig. 220 is believed to be qualitatively correct and essentially accurate quantitatively. There is somewhat more uncertainty with respect to the ternary iron-chromium-carbon diagram, principally because the sluggishness of the reactions makes it difficult to establish definite transformation temperatures, and because of the formation of complex carbides containing the three components in varying proportions. From a practical standpoint the limits of the gamma and gamma plus alpha regions are of greatest significance and these have been established with reasonable accuracy. Sections from the ternary diagram taken at constant carbon or constant chromium contents are convenient for studying these limits. Diagrams for constant chromium contents of 8, 12, 15, and 20% are given in Figs. 221 to 224. (Based on the reports of Tofaute and coworkers.[9] See also " Alloys of Iron and Chromium," Vol. I.[x])

Comparing these diagrams with the iron-iron carbide diagram (Fig. 84, p. 158), the following trends are noted. Chromium reduces the limits of the single phase gamma field, which disappears entirely at about 20% chromium, and increases the extent of the alpha and alpha plus gamma fields. In commercial alloys containing appreciable amounts of manganese, copper, and other elements, the phase boundaries are probably altered somewhat. For example, there is evidence that the boundary between the alpha and alpha plus gamma fields should be shifted farther to the left in Figs. 223 and 224.

The commercial iron-chromium-carbon stainless steels are produced in several grades ranging from hard non-shrinking, non-deforming tool steels to soft ferritic irons suitable for drawing and forming. The hardenable grades may be classified as martensitic steels and all others as ferritic steels, based on the microstructures of their matrices.

Martensitic Stainless Steels — High Carbon. — In order to quench-harden any steel it is necessary that its structure be essentially austenitic at the quenching temperature, therefore the compo-

sition range available for full hardening by transformation of austenite to martensite is evidently reduced by chromium.   On the other hand, the sluggishness of the transformation of chromium-rich austenite reduces the critical cooling rate very drastically, making it possible to obtain full hardness upon air or oil quenching. Alloys having compositions similar to those produced by Brearley fall in the hardenable range and after quenching their microstructures are similar to those of other hardened steels.   They may likewise be tempered to obtain various combinations of mechanical properties.   In order, however, to develop their stainless characteristics to the fullest extent, the carbon and chromium must be retained in solid solution.   According to Fig. 222, under conditions of equilibrium below the critical transformation range (as approached by cooling slowly from high temperatures or by tempering hardened structures) these alloys consist of complex carbides in an alpha matrix.   It has been found that chromium existing as carbides does not contribute to the protection of the matrix and may, in fact, accelerate corrosion, therefore it is necessary to dissolve the carbides at a high temperature and retain them in solution by quenching.

Obviously the formation of carbides depends on the presence of carbon in the alloy; hence it should be possible to secure adequate protection from the chromium in very low carbon alloys without resorting to quenching.   It was not possible to produce such low carbon stainless steels on a commercial scale until about 1920 because low carbon ferrochromium additions were not yet available, and carbon could not be removed satisfactorily in the melting furnaces without excessive loss of chromium.   Furthermore, the early applications were chiefly stainless cutlery, for which the high hardness of the martensitic structure was a necessity.   This class of stainless steels is now a minor part of the total tonnage, most of which is made up of low carbon alloys.

In practice the cutlery compositions now range from about 12% chromium with 0.30 to 0.40% carbon to 18% chromium with 0.70 to 1.0% carbon (Type Nos. 420 and 440).*   The "original" cutlery grades contain the lower carbon and chromium contents, while the later modifications are higher in both carbon and

* See page 477 for base prices and standard compositions.

chromium, giving greater hardness and superior cutting qualities. The 1.0% carbon type is used for ball bearings, pump and valve parts, and equipment subjected to erosion and abrasion as well as corrosion.

Comparing Figs. 222 to 224, it is evident that as the chromium content is increased, the carbon content must also be increased in order to avoid the presence of untransformed alpha phase at heat treating temperatures. To obtain a uniformly and fully hardened steel on cooling, the structure at the quenching temperature must be austenitic or a mixture of saturated austenite and excess carbides. In addition to using a suitable composition, the heat treatment must be carefully performed, using a hardening temperature range of 1800°–1850° F. (980°–1000° C.) for the original cutlery grade and about 50° F. higher for the modified or higher chromium grades. Because of their relatively low thermal conductivity, the heating time should be longer for stainless than for ordinary steels. Preheating or soaking at a lower temperature, such as 1450° F., is helpful in treating large or irregular sections. While most sections harden by air cooling, oil quenching is generally recommended for uniformity and better mechanical properties. Water quenching is unnecessary and may cause cracking. Tempering temperatures up to about 900° F. are used to relieve quenching stresses. The high chromium martensites are more resistant to decomposition and softening than ordinary martensite, nevertheless high tempering temperatures are to be avoided if full wear and corrosion resistance are desired.

In order to soften these steels for machining purposes they are heated to 1600° F. and cooled slowly at a rate of not more than 50° F. drop per hour. (See Table LXV for typical properties.)

The smoother the surface finish can be made the greater will be the resistance to staining and corrosion. While this is true of all grades of stainless, and most other alloys, a bright finish is especially important when using the martensitic grades. It is also necessary that all traces of scale and all foreign substances be removed from the surface to prevent local areas of attack. Scale from rolling or heat treating is removed by a rough grinding or pickling

operation. A final treatment in strong nitric acid is sometimes applied to insure removal of particles of iron or other substances which may have become embedded in the surface, and also to passivate the steel. (See page 469.)

The mechanical properties of all grades of stainless steels, as well as detailed information on compositions, processing operations, heat treatments, corrosion resistance, oxidation resistance, and applications, are now available in a number of different sources, in-

TABLE LXV.   TYPICAL PROPERTIES OF CUTLERY GRADES OF
STAINLESS STEELS

	0.35% C–13.0% Cr *		0.70% C–17.0% Cr **	
	Hardened	Annealed	Hardened	Annealed
Tensile strength, lb./sq. in. . .	240,000	100,000	270,000	95,000
Yield strength, lb./sq. in.	200,000	65,000	245,000	54,000
Reduction of area, %	8	60	3.5	45
Elongation in 2 inches, %	4	27	2	27
Brinell hardness	500	185	545	185
Rockwell hardness	52 " C "	91 " B "	55 " C "	91 " B "

* Quenched in oil from 1825° F., tempered at 450° F. Annealed at 1600° F.
** Quenched in oil from 1850° F., tempered at 450° F. Annealed at 1650° F.

Courtesy of Owen K. Parmiter, Metals Handbook.[12]

cluding " The Book of Stainless Steels "[10], " Stainless Iron and Steel "[11] (British practice), " Metals Handbook," " The Alloys of Iron and Chromium," Vol. II[8], and the literature of the manufacturers of these steels. The specific data included in this text are intended to indicate the general characteristics of the basic grades.

Martensitic Stainless Steels — Low Carbon. — The high carbon martensitic stainless steels have mechanical characteristics similar to those of tool steels. Low carbon grades are more nearly comparable to the S.A.E. machine steels in their properties and applications. The nominal composition of the most important low carbon grade is 10.0 to 14.0% chromium with under 0.15% carbon (Type No. 410). A " turbine quality " grade (Type No. 403) is also within this composition range.

Despite the low carbon content of Type No. 410, tensile strengths of 180,000 lb. per sq. in. are obtained after quenching and drawing

at temperatures up to about 800° F. At higher drawing temperatures excellent combinations of strength, toughness, and ductility are possible as shown in Fig. 225.

From Fig. 222 it is apparent that a 0.1% carbon, 12.0% chromium steel is austenitic at 1800° F., its quenching temperature. Higher chromium contents tend to give mixed austenite-ferrite structures and, if the carbon content is sufficiently low, austenite-
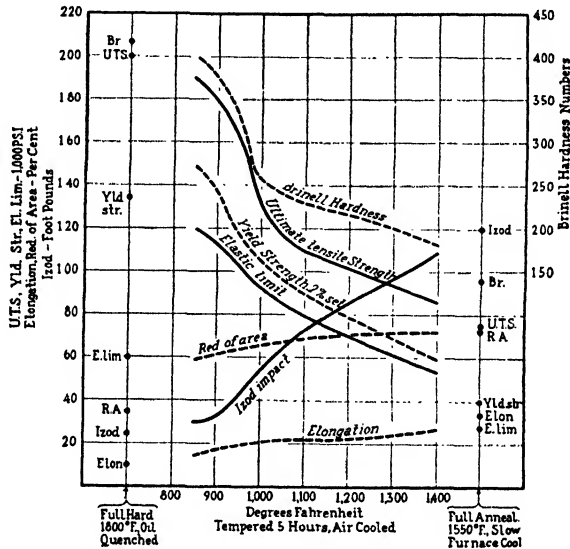
FIG. 225. — TYPICAL PROPERTIES OF 0.11% CARBON, 12.5% CHROMIUM STEEL HEAT TREATED BARS 1 INCH IN DIAMETER. (*Courtesy of Rustless Iron and Steel Corporation.*)

free structures; therefore it is necessary to keep the composition properly balanced to retain hardenability.

As in the case of the cutlery grades, corrosion resistance is reduced by tempering at high temperatures; however, the reduction is not nearly as marked as in the higher carbon steels, and tempering temperatures over 900° F. are often used to obtain the mechanical properties required for machine parts and equipment subjected to abrasion and mild corrosion in the coal, paper, foundry, mining, and other industries.

Free-Machining Low Carbon Martensitic Stainless Steel. — The stainless steels as a class are considered to be the most difficult

TABLE LXVI. COMPOSITIONS AND BASE PRICES OF STAINLESS STEELS

(Standard Type Numbers and Analyses — Revised to March 1, 1939)

Type No.	Carbon	Chromium	Nickel	Other Elements	Base Price in Cents/lb. Second Quarter 1939		
					Bars, Drawn Wire, Structurals	Cold Rolled Strip	Sheets
301	0.09–0.20	16.0–18.0	7.0– 9.0	Mn 1.25 Max.	24.0	25.5	34.0
302	0.08–0.20	18.0–20.0	8.0–10.0	Mn 1.25 Max.	24.0	28.0	34.0
302B	Over 0.08	18.0–20.0	8.0–10.0	Si 2.0–3.0 Mn 1.25 Max.	24.0	28.0	34.0
303	0.20 Max.	18.0–20.0	8.0–10.0	S or Se 0.07 Min. or Mo 0.60 Max.	26.0	33.0	36.0
304	0.08 Max.	18.0–20.0	8.0–10.0	Mn 2.0 Max.	25.0	30.0	36.0
308	0.08 Max.	19.0–22.0	10.0–12.0	Mn 2.0 Max.	29.0	35.0	41.0
309	0.20 Max.	22.0–26.0	12.0–14.0		36.0	47.0	47.0
309S	0.08 Max.	22.0–26.0	12.0–14.0		40.0	51.0	51.0
310	0.25 Max.	24.0–26.0	19.0–21.0		49.0	56.0	53.0
316	0.10 Max.	16.0–18.0	14.0 Max.	Mo 2.0–3.0	40.0	48.0	48.0
321	0.10 Max.	17.0–20.0	7.0–10.0	Ti Min. 4 × C	29.0	38.0	41.0
329	0.10 Max.	25.0–30.0	3.0– 5.0	Mo 1.0–1.5	36.0	45.0	44.0
330	0.25 Max.	14.0–16.0	33.0–36.0		49.0	—	53.0
347	0.10 Max.	17.0–20.0	8.0–12.0	Cb 10 × C	33.0	42.0	45.0
403	0.15 Max.	11.5–13.0	——	Turbine Quality	21.5	27.0	29.5
410	0.15 Max.	10.0–14.0	——		18.5	22.0	26.5
416	0.15 Max.	12.0–14.0	——	S or Se 0.07 Min. or Mo 0.60 Max.	19.0	23.5	27.0
420	Over 0.15	12.0–14.0	——		24.0	36.5	33.5
420F	Over 0.15	12.0–14.0	——	S or Se 0.07 Min. or Mo 0.60 Max.	24.5	—	—
430	0.12 Max.	14.0–18.0	——		19.0	22.5	29.0
430F	0.12 Max.	14.0–18.0	——	S or Se 0.07 Min. or Mo 0.60 Max.	19.5	24.5	29.5
431	0.15 Max.	14.0–18.0	2.0 Max.		19.0	22.5	29.0
440	Over 0.12	14.0–18.0	——		24.0	36.5	33.5
441	Over 0.15	14.0–18.0	2.0 Max.		24.0	36.5	33.5
442	0.35 Max.	18.0–23.0	——		22.5	32.0	32.5
446	0.35 Max.	23.0–30.0	——		27.5	52.0	36.5
501	Over 0.10	4.0– 6.0	——		8.0	17.0	15.75
502	0.10 Max.	4.0– 6.0	——		9.0	18.0	16.75

(About 12 additional types are recognized by most producers.)

Note: The carbon ranges of Types 301, 302, 302B, and 303 may be specified to a four-point range within the above limits. When the carbon content of Types 301, 302, or 302B is specified as 0.11% or under, the price of Type 304 applies.

Both compositions and base prices are subject to change.

of all commercial steels to machine. While the various grades differ in their machinability, all but the special free-cutting compositions have the characteristic of sticking to the cutting tool or seizing due to high frictional qualities.[13] Free-machining characteristics are imparted to the low carbon martensitic steel by incorporating molybdenum and sulphur or phosphorus and selenium (Type No. 416). The toughness, ductility, and corrosion resistance are reduced by high sulphur contents, the hot working characteristics are poor, and welding is not recommended; however, greatly improved machinability and non-seizing characteristics are obtained with only slight reduction in strength compared with Type No. 410. Additions of phosphorus and selenium for free-machining purposes have less influence on the corrosion resistance than sulphur.

Applications of Type No. 416 include screw machine products, valve parts, pump shafts, and wearing surfaces operating under high pressures or at elevated temperatures. The normal tendency of stainless steels to seize under these conditions is overcome by the non-galling properties of the free-machining grade. Heat treating mating parts to different hardnesses also reduces seizing.

Ferritic Stainless Steels. — Examination of Figs. 223 and 224 shows that low carbon and high chromium contents give steels which may be heated to any temperature up to their melting points without transformation to austenite. The freedom from quench hardening which is thus insured is of great value in the manufacture and fabrication of these steels. On the other hand, the absence of the alpha to gamma transformation eliminates the possibility of grain refinement by heat treatment. As in the case of many other metals and solid solution alloys, the processes of cold work and recrystallization may be used to control grain size and mechanical properties. Recrystallizing temperatures up to about 1550° F. may be used without excessive grain growth of heavily cold worked material. Although the ductility is restored by a short anneal at 1400° F., temperatures of 1550° F. and higher may be used for full annealing.

The commercial ferritic stainless steels range from about 14 to 30% in chromium content with 0.05 to 0.35% carbon. The higher carbon contents are permissible only in the very high chromium

wrought alloys. Even 0.10% carbon is sufficient to permit considerable quench hardening in commercial alloys in the range 14 to 16% chromium.

17% Chromium Ferritic Stainless Steel. — The principal ferritic grade is Type No. 430 whose nominal range is 14 to 18% chromium with under 0.12% carbon. Steels of this type ordinarily contain about 16.5 to 17.5% chromium and under 0.10% carbon. Although subject to quench hardening, especially when carbon is near the upper limit, this grade is not ordinarily heat treated for hardness, therefore it is properly included in the ferritic classification. The lower carbon content and the comparatively small amount of chromium present as carbides account for the superior corrosion resistance of ferritic compared with martensitic steels with the same chromium content. The forming and drawing characteristics of the ferritic stainless steels approach those of mild steel. In the form of sheets, bars, and wire, Type No. 430 is widely used for architectural work, automotive trim, hardware, and equipment in the chemical and food industries.

FIG. 226. — FERRITIC STAINLESS STEEL CONTAINING 17.2% CHROMIUM, 0.07% CARBON. (Etched by electrolysis using 1% HCl in methyl alcohol.) X100. (*Courtesy of W. L. Havekotte.*)

The microstructure of a sheet steel containing 17.2% chromium and 0.07% carbon is shown in Fig. 226. The fine ferrite grain size is the result of a 50% cold reduction followed by annealing at 1500° F. Very small carbide particles appear within the grains.

Type No. 430 is also resistant to heavy scaling in furnace atmospheres at temperatures up to about 1550° F. The thin oxide film which forms is tightly adhering and gives protection against further oxidation. Because of this behavior, the 17% chromium steel has

many applications in processing equipment, furnace parts, oil burners, heat exchangers, and other equipment operating at elevated temperatures. It is superior to the iron-chromium-nickel heat resisting alloys (see page 506) in resisting sulphur-containing gases. The mechanical properties of this grade are included in the first group of Table LXVII.[14]

As in the case of the lower chromium martensitic grade (10 to 14% Cr), free-machining properties may be obtained by the addition of molybdenum and sulphur, or phosphorus and selenium, giving Type No. 430F.

27% Chromium Ferritic Stainless Steel. —The temperatures to which iron-chromium alloys may be heated without heavy oxidation or scaling increase with the chromium content. For applications involving temperatures between 1500° and 2100° F., Type No. 446 is the most suitable of the ferritic alloys. (Type No. 442 is also used for temperatures up to 1800° F.) The nominal composition range of Type No. 446 is 23 to 30% chromium with under 0.35% carbon, and the most widely used composition is about 27% chromium with 0.10 to 0.20% carbon. The microstructure of this steel consists of a ferritic matrix in which complex iron-chromium carbides are embedded. Although the 27% chromium steel generally has a higher carbon content than the 17% chromium ferritic grade and a considerable amount of chromium is, therefore, " tied-up " in the form of carbides, the chromium content in solution in the matrix is sufficient to give it excellent resistance to oxidation at high temperatures, as well as high resistance to nitric acid and to many other corrosive substances of an oxidizing nature.

This grade cannot be hardened nor can its grain size be refined by heat treatment. When the grain size is coarse after rolling or forging or becomes coarse after exposure to very high temperatures (approximately 2000° F.) the ductility drops far below the normal values reported in Table LXVII. In the coarse grained condition it is difficult to carry out cold forming, drawing, and other fabricating operations. The use of high nitrogen-bearing ferrochromium in making these steels greatly reduces their tendency towards coarse grain size.[15]

Although the static ductility of this grade is satisfactory when

TABLE LXVII. CHARACTERISTICS OF FERRITIC IRON-CHROMIUM-CARBON-ALLOYS

	16 to 20% Cr (0.12 C max.)		23 to 30% Cr (0.30 C max.)	
Microstructure	Ferrite with a little pearlite		Ferrite	
Effect of heat treatment	Somewhat modified		Not modified	
Effect of cold work	Modified		Modified *	
	Annealed	Cold Worked	Annealed	Cold Worked
Room temperature properties				
Tensile strength, lb/sq. in.	75,000 to 90,000	90,000 to 190,000	75,000 to 95,000	85,000 to 175,000
Yield point, lb/sq. in.	40,000 to 55,000	65,000 to 130,000	45,000 to 60,000	55,000 to 135,000
Elongation in 2 in., %	30 to 20	20 to 2	30 to 20	25 to 2
Reduction of area, %	55 to 40	40 to 20	60 to 50	50 to 20
Izod impact, ft. lb.	75 to 5	30 to 2	Very low	Very low
Brinell hardness	140 to 180	175 to 275	140 to 180	150 to 250
Strength when hot				
1000° F.	47,500 to 54,000		49,000 to 52,000	
1100° F.	37,000 to 41,000		20,000 to 34,000	
1300° F.	15,000 to 23,000		10,000 to 15,000	
1500° F.	8,000 to 12,000		6,000 to 7,500	
Creep strength (1% in 10,000 hr.)				
1000° F.	5,500 to 8,500		6,000 to 7,000	
1100° F.	2,300 to 5,000		2,700 to 3,300	
1200° F.	1,300 to 2,100		1,600 to 1,800	
1300° F.	1,000 to 1,400		500 to 800	

* Age hardening and similar treatments are excluded.

Courtesy of V. N. Krivobok, Metal Progress.[14]

the grain size is not coarsened excessively, it is quite brittle at room temperature under impact loading, a characteristic which should be considered in designing and using equipment made from this material.

The possibilities for fabricating the ferritic stainless steels by forming, drawing, stamping, and other processes are reflected in their mechanical properties; however, the superior workability of the 17% chromium grade is not evident from the properties given in Table LXVII. The 27% chromium steel strain-hardens more rapidly and does not lend itself readily to cold fabrication operations requiring severe forming or drawing. It may be softened after cold working by annealing at 1450° to 1650° F. followed by air cooling.

The alloys containing about 20% chromium and over are especially subject to loss of ductility at room temperature if previously held for any appreciable time, either during heat treatment or in service, at temperatures in the neighborhood of 900° F. or 475° C., therefore rapid cooling from the annealing temperature is recommended. Heating to somewhat higher temperatures followed by rapid cooling completely restores the ductility, which is a strong indication that the embrittlement is caused by precipitation of a second constituent from the ferrite. Becket [16] calls this phenomenon the " 475° brittleness " and discusses the possible causes, including precipitation of the sigma phase. In Fig. 220 this phase is shown at 50 atomic per cent chromium, corresponding to the compound FeCr. The extent to which this phase forms mixtures with the alpha solid solution has not been fully determined. The boundaries of the two-phase regions are indicated tentatively in Fig. 220; however, more recent observations indicate that the left hand boundary may extend to lower chromium contents. Because of the varying solubility of chromium in the alpha phase (indicated by the slope of the alpha-sigma boundary line) the possibility of precipitation hardening is clearly indicated; nevertheless, Becket and his co-workers were not able to link the 475° brittleness directly with the sigma phase and concluded that some other phase of unknown composition is responsible for this effect.

Jette and Foote [17] have shown that silicon and certain other ele-

ments, as well as cold working, tend to accelerate the formation of the sigma phase.

Krivobok [18] has also investigated age hardening and brittleness of the high chromium-iron alloys in considerable detail.

Welding of the Iron-Chromium Stainless Steels. — The strong air hardening tendency of the martensitic stainless steels makes welding a very difficult problem (except in very low carbon grades) involving preheating and annealing before the metal cools from the welding heat in order to prevent cracking. However, by means of an annealing treatment after welding the grain size may be refined and the metal softened and toughened. The free-machining modifications of both the martensitic and ferritic steels have poor welding characteristics.

The ferritic steels are subject to grain coarsening at the weld, in which case ductility and impact toughness are reduced. The 27% chromium type is especially susceptible to coarsening and, of course, cannot be refined by heat treatment. High nitrogen additions are particularly beneficial in high chromium welding rods and steels to be welded. According to Franks [15] the nitrogen is practically all recovered in the weld metal, giving fine grained deposits of good ductility for this class of steels.

In welding both martensitic and ferritic stainless steels best results are obtained with iron-chromium-nickel filler rods.[14] The 18% chromium-8% nickel and the 25% chromium-12% nickel types are often used. A considerable part of the nickel leaves the deposited weld metal and penetrates well into the adjacent parent metal, improving the toughness and ductility.

In welding the 17% chromium ferritic grade air hardening may occur, although to a much smaller extent than in the martensitic steels. Hardening from this source may be relieved by annealing.

Effect of Alloy Additions on Iron-Chromium-Carbon Steels. — The basic compositions of the stainless steels may be modified by additions of special elements. An entirely new series of alloys results upon the addition of 7% or more nickel. Large manganese additions have a similar drastic effect on the properties and structures of stainless steels. In this section much smaller alloy additions will be considered.

Nickel, even in small amounts, exerts its usual influence as an austenite former and stabilizer. Nickel contents up to 2% are used in both the martensitic and the normally ferritic 17% chromium grade to increase air hardening capacity, strength, and elastic ratio (Type Nos. 441 and 431 respectively). An 18% chromium-2% nickel steel is popular in England for engineering applications requiring the hardness and strength of the lower chromium martensitic grades together with superior corrosion resistance.[19] Among its uses are steam and hydraulic fittings and naval aircraft parts.

It is quite apparent that the 17% chromium grade, No. 430, is a border line alloy, capable of hardening by heat treatment when the chromium is on the low or the carbon on the high side of the permissible range, or when austenite forming additions such as nickel or copper are made. On the other hand, low carbon, high chromium, or the addition of ferrite stabilizing elements such as titanium [20, 21], columbium [21], and aluminum [14] tend to reduce its hardenability. Titanium and columbium are strong carbide formers and stabilizers and there is evidence that columbium carbides (and to a lesser extent titanium carbides) are difficult to dissolve at high heat treating temperatures. (See the discussion of a paper by Franks.[22]) Since undissolved carbides reduce the carbon available in the austenite, or, perhaps, drastically reduce the amount of austenite formed, the reduced hardenability is to be expected. Krivobok [14] has shown that small additions of aluminum also reduce air hardening very appreciably, thereby improving the ductility of welded parts.

Silicon and aluminum reduce scaling at elevated temperatures when added to the heat resisting stainless steels. Silicon contents of 2 to 3% are common in several grades of chromium-nickel austenitic as well as in straight chromium steels.

The iron-chromium-silicon phase diagram has been investigated by Anderson and Jette.[23] Constitutional diagrams for Fe-Cr-C-Si alloys and information on the use of silicon in the 27% chromium grade are reported by Poboril.[24]

Aluminum additions up to about 5% are made, especially in foreign practice, to improve the heat resistance of several straight chromium compositions. An alloy containing 37.5% chromium,

7.5% aluminum, and 0.15% maximum carbon is said to withstand continuous use at temperatures up to 2375° F.[14]

Tungsten and molybdenum additions to certain heat resisting grades increase their strength at elevated temperatures.

The use of sulphur, selenium, and molybdenum for improving machinability and of nitrogen for grain size control of ferritic grades has already been mentioned.

The essential characteristics of martensitic and ferritic stainless steels are included in a very convenient comparative summary arranged by F. R. Palmer [25] and reproduced through the courtesy of the American Society for Metals as Table LXVIII.

Intermediate Chromium Steels. — In Chapter IX reference was made to a 5% chromium steel used for equipment in the oil refining industry. This composition is the most important of a group of steels containing 5 to 9% chromium with small addition of molybdenum, tungsten, silicon, aluminum, titanium, columbium, and other elements. In composition, cost, and corrosion and heat resistance they are intermediate between non-alloyed steels and the stainless types. Discussion of these materials has been included here despite the fact that they do not qualify as high alloy steels according to the arbitrary limit of 12% given at the beginning of this chapter.

The metal temperatures in oil stills are normally below 1200° F., therefore the use of high alloy stainless steels is not warranted from the standpoint of resistance to scaling. Furthermore, the intermediate alloy compositions withstand the corrosive effects of high sulphur crude oils for reasonably long service periods and have sufficiently high strength at the operating temperatures to give a long life before removal is necessary to avoid creep failures, thus the 5% chromium steel has been found to be economical for still tubes, heat exchanger tubes, header castings, piping, and valves in the oil industry. The conditions of temperature and pressure under which certain equipment is operated in the power industry are quite similar and the intermediate alloy steels have been adopted, for example, for steam superheater tubes and headers.

The low carbon content of Type No. 501 reduces somewhat the natural air hardening tendency of these steels. In the annealed state it has mechanical properties similar to those of mild steels and

TABLE LXVIII.   THE A-B-C OF CORROSION AND HEAT RESISTING
STEELS

Arrangement due to F. R. Palmer.[25]   Adapted from " The Book of Stainless Steels," Second Edition.   Courtesy of the American Society for Metals.

GROUP A (MARTENSITIC)	GROUP B (FERRITIC)	GROUP C (AUSTENITIC)
CHEMICAL ANALYSIS Chromium less than about 16%: carbon less than about 0.40%. May contain small percentages of tungsten, copper, nickel, silicon, columbium, aluminum, and more frequently molybdenum. Group is magnetic.	Chromium more than about 16%; carbon quite low, but can increase as chromium goes up. May contain small percentages of copper, nickel, silicon, molybdenum, tungsten, nitrogen. This group is magnetic.	Contains enough nickel to make steel austenitic and non-magnetic. They usually contain twice as much chromium as nickel or vice versa; total alloy content at least 26%. Carbon quite low. (Manganese sometimes substituted for nickel, in part.)
HEAT TREATMENT Respond to hardening and tempering. Resulting physical properties depend on chemical analysis (principally carbon content).	Do not respond. 18% chromium toughened by long anneal at more than 1400° F., and air cooling. Avoid decarburizing the skin. 25% chromium gets best strength and toughness by rapid cooling from 1650°.	Do not respond to hardening by heat treatment. Must be rapidly cooled from soaking heat at 1900° to 2150° F. to retain austenitic structure free of carbides. (Brinell 140 to 170.)
TOUGHNESS Are structurally dependable. After tempering are not brittle in notched sections or under impact.	Laminated structure, from coarse ferrite in ingot, causes low impact values, but proper rolling and heating gives adequate toughness in rods, bars, and sheets. Structure is refined by nitrogen.	Extremely tough at all temperatures down to liquid air. Dependable against shock except when corroded at grain boundaries (a preventable condition).
GRAIN GROWTH AND STRUCTURAL CHANGES AT HIGH TEMPERATURES Not subject to excessive grain growth. Thoroughly dependable for supporting any load or shock within their carrying capacity up to 1400° F. Brittleness in plain chromium steels when cooled after long heating is avoided by addition of molybdenum.	The chromium-irons low in carbon and those high in silicon or aluminum (when cold worked) are subject to excessive grain growth, especially above 1900° F. Grain growth reduced by nitrogen. Long service at 800° to 950° F. makes them brittle when cold, although they are not brittle at working temperatures.	Alloys near the austenite-martensite border line tend to precipitate carbides at grain boundaries during service at 800° to 1600° F., losing some toughness and becoming susceptible to intergranular attack. This is controlled by very low carbon, by titanium or columbium, by increasing the chromium and nickel, or by prior "stabilization."
STRENGTH AT ELEVATED TEMPERATURES Much better than straight carbon steel for temperatures up to 1000° or 1200° F. Retain tensile properties up to 750° F.	Heat resisting varieties quite tough at temperatures up to 1600° F. Superior in ductility to Group C but not in creep resistance.	Have high creep strength up to 1200° F. which is enhanced by tungsten or molybdenum. Toughness impaired in nonstabilized alloys by service at 800° to 1600° F.

GROUP A (MARTENSITIC)	GROUP B (FERRITIC)	GROUP C (AUSTENITIC)
HOT WORKING QUALITIES Readily forged, pierced, or rolled at 2000° to 1700° F. Preheat and soak stock at 1600° F. Plain chromium alloys air harden on cooling.	May be forged, rolled, or pierced. Should be heated quickly. Forge from 2200° F. down to 1750° F. On last heat continue cold working to 1400° F. to refine grain. Alloys do not air harden.	May be forged, rolled, or pierced. Preheat and soak at 1600° F., heat quickly to 2200° F., forge down to 1850° F. Hot short range: 1800° to 1300° F. Alloys do not air harden.
COLD WORKING QUALITIES Low carbon varieties can be easily cold drawn into wire, cold rolled, bent, formed, upset, coined, and deep drawn.	Can be cold drawn into wire, cold rolled, bent, formed, upset, coined, and deep drawn, especially when warm (300° to 500° F.).	Can be cold drawn into wire, cold rolled, bent, formed, upset, coined, and deep drawn. Work-harden twice as rapidly as Groups A and B.
MACHINABILITY Machine satisfactorily with properly designed tools when heat treated to 200 to 250 Brinell. Free-cutting grade contains complex sulphides or selenium.	Machine satisfactorily with properly designed tools. Cold working and high sulphur or selenium improve machinability.	Most difficult of all even with super high speed and carbide tools. Use sharp tools having greater top rake than usual, and cut continuously. Free-cutting grade contains selenium and phosphorus.
RIVETING Make excellent cold rivets. Air hardening, plain steels not recommended for hot rivets driven above 1500° F.	Extra precautions required to avoid brittle rivets. Conical heads should be cold upset on ground bars; rivets driven at 1425° F. into chamfered holes.	Excellent for either hot or cold rivets. Hot rivets may be driven at a high temperature (1900° F.).
WELDING PROPERTIES Preheated parts can be welded with gas, electric arc, or resistance. Anneal immediately before weld air hardens. Little grain growth. Columbium or aluminum increases ductility of welds.	Can be welded. Anneal at 1450° F. to reduce embrittlement alongside weld. Those metals subject to grain growth are brittle adjacent to the weld. Those metals not subject to grain growth yield satisfactory welds.	Can be welded with gas, electric arc, or resistance, if carburization is avoided. Weld does not air harden and is very tough. Only the relatively low carbon or "stabilized" metal should be welded if article must resist corroding media.
CORROSION RESISTANCE Increases with chromium content; inferior to Groups B and C. Resists weather, water, steam, and mild corrodents when chromium is 11.5% or more. If carbon is relatively high, metal must be hardened and tempered (below 1000° F.).	Possess corrosion resisting properties superior to Group A; increases with chromium content. Especially good for nitric and other oxidizing acids.	Corrosion resistance depends largely upon total alloy content. Resists nearly all corrodents measurably better than Groups A and B; especially good for organic acids. Severe pitting may occur in stagnant chloride solutions under particles of foreign matter and along faying surfaces.
SCALE RESISTANCE Increases with chromium content. Generally useful for continuous temperatures up to 1200° F., and in some services up to 1500° F.	Superior to Group A, especially when chromium is about 25%; then resist reducing atmospheres up to 2100° F., oxidizing up to melting points, and sulphur gases up to 1800° F.	Excellent where combination of high temperature and corrosion is to be met. High chromium, low nickel alloys required to resist sulphurous gases. Addition of 2 to 3% silicon markedly improves scale resistance.

may, therefore, be used where cold forming or other cold working operations are applied in fabrication, and for parts to be welded. Air hardening after welding may be eliminated by suitable additions of titanium (5 to 8 times the carbon) or columbium.[22]

The higher carbon type, No. 502, air hardens very readily, developing tensile and impact properties comparable to those of heat treated alloy machine steels. Air hardened and tempered steels may be used for bolts, valves, and other parts operating at temperatures lower than the tempering temperature used in heat treating; however, fully annealed steels are used for operation at higher temperatures.

Unexpected brittle failures occurred in early installations of these steels in oil stills. It was found that after cooling from the operating temperatures the metal had very low impact toughness and was subject to cracking during removal of the coke deposited within the tubes. The brittle condition was entirely absent upon reheating to the working temperature. Although failures by cracking were reduced in number by altering the design and methods of cleaning the tubes, the addition of 0.5% molybdenum to the steel solved the problem by eliminating room temperature brittleness. Steels containing 1% tungsten have also given excellent service.

It is interesting to note that during the investigation of this problem Wilten[26] observed a direct correlation between brittle failures in service and the occurrence of temper brittleness. Susceptibility to temper brittleness was measured by the ratio of the Charpy impact value obtained by cooling slowly to that obtained by quenching from a tempering temperature of 1200° F. (Specimens previously hardened by quenching in water from 1650° F.) In discussing Wilten's paper, Malcolm and Jones review the subject of temper brittleness, pointing out the inconsistencies in data reported in the technical literature as well as in correlation of susceptibility to temper brittleness with service results. The uncertainties associated with temper brittleness arise from the varying response, from heat to heat, of compositions which are likely to be brittle, and from the uncertainties of the impact test itself. Nevertheless, it is generally agreed that the trends indicated by intelligent application of both the impact

tests and the temper brittleness susceptibility tests are quite definite and yield valuable information on the serviceability of steels.

Tube failures not involving brittleness are generally caused by a combination of oxidation and creep.[28] Molybdenum and tungsten additions increase the creep strength as shown in Fig. 227. At 900° F. the stress required to produce 1% elongation in 10,000 hours in the 5% Cr-0.5% Mo steel was nearly twice as high as in the case of the 5% Cr steel and three times as high as in the low carbon steel.

Modifications containing silicon and aluminum have been found to reduce scaling. Steels containing rather high chromium and silicon contents have been used for many years for automotive and aircraft engine valves and similar high temperature applications. More recently silicon and combinations of silicon and aluminum have been

FIG. 227. — CREEP STRESSES FOR METALS USED IN OIL REFINERY TUBES. A COMPOSITE OF WORK AT THREE LABORATORIES, CARRIED OUT FOR AT LEAST 5000 HOURS. (*Courtesy of E. C. Wright, The Book of Stainless Steels.*[27])

added to tonnage steels of the type under discussion. White, Clark, and McCollam [29] investigated various combinations of chromium, silicon, and aluminum in the intermediate chromium range and found that additions of both silicon and aluminum were much more effective than equal chromium additions in imparting oxidation resistance.

Chromium contents higher than 5% and approaching the 12% chromium stainless steels are useful under severe operating conditions in which high creep strength and resistance to oxidation and corrosion are needed. Newell [30] has described the development of a 9% chromium-1.25% molybdenum steel for use at temperatures up to 1250° F. Surface oxidation in this temperature range decreases with increasing chromium content as shown by Newell in the following data based on weight losses in 250 hours in an oxidizing atmosphere at 1200° F.:

Relative weight loss

Carbon steel 100
2% Cr, 0.50% Mo 37
5% Cr, 0.50% Mo 14.8
9% Cr, 1.50% Mo 5.5
12% to 14% Cr 1.8
18% Cr, 8% Ni 0.0

Silcrome Valve Steels. — The iron-chromium-silicon valve steels, mentioned above, have never been considered to be stainless alloys; however, they are similar in composition to the intermediate chromium steels. A typical alloy for gasoline engine exhaust valves contains 8.5% Cr, 3.25% Si, 1.5% W, and 0.45% C.[31] Materials of this type were giving excellent service as valve metals long before the tonnage types of intermediate chromium alloys were developed.

Austenitic Stainless Steels. — Nearly one-half of all the wrought stainless steel now being produced is of the well known 18% chromium-8% nickel (or simply 18–8) composition, including the modifications represented by Types 301 to 308 and Types 316, 321, and 347. Despite the increasing importance of modified compositions the basic type, No. 302, is by far the leader in tonnage produced. The higher alloy austenitic steels, Types 309, 309S, and 310, are essentially heat resisting alloys.

Although nickel extends the range of exposure conditions in which stainless steels remain passive, the principal justification for using the more costly austenitic rather than the ferritic grades is their greater ease of fabrication, including cold drawing or forming operations and welding. The austenitic grades also surpass the ferritic in creep strength at elevated temperatures and in toughness at normal and low temperatures down to that of liquid air. Other general characteristics are compared with those of the ferritic and martensitic groups in Table LXVIII, page 486.

The manner in which chromium reduces and finally eliminates the single phase gamma region of the equilibrium diagram was shown in Figs. 221 to 224. Krivobok and Grossmann[32] in a similar series of diagrams illustrate the manner in which nickel restores the gamma region of steels containing 18% chromium and variable carbon contents. Aborn and Bain[33] have outlined the phase regions of a por-

tion of the ternary iron-nickel-chromium diagram as shown in Fig. 228. This diagram applies to very low carbon alloys cooled rapidly from temperatures at which austenite extends over the widest composition range. These temperatures vary with the composition, therefore Fig. 228 is not strictly an equilibrium diagram based on a fixed temperature (isothermal section) but represents the ap-
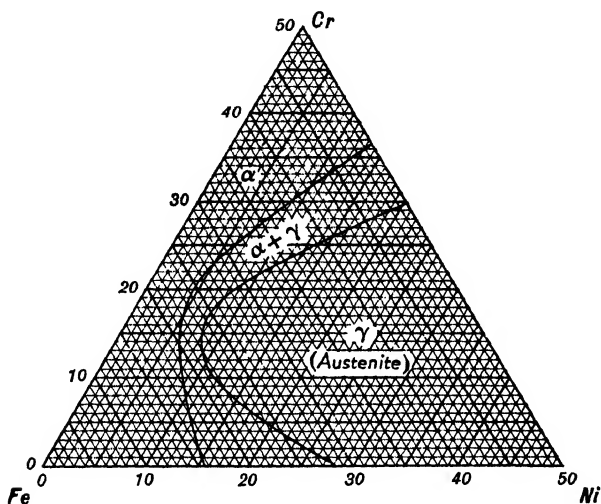
FIG. 228. — APPROXIMATE PHASE RELATIONS OF IRON-CHROMIUM-NICKEL ALLOYS COOLED RAPIDLY FROM TEMPERATURES AT WHICH AUSTENITE EXTENDS OVER THE WIDEST COMPOSITION RANGE. (*Courtesy of R. H. Aborn and E. G. Bain, Trans. Am. Soc. for Steel Treating.*[33])

proximate phase relations of alloys cooled rapidly from high temperatures. The practical importance of this diagram lies in the fact that commercial " annealing " treatments for these alloys also consist of quenching from high temperatures in order to establish the austenitic phase and prevent its decomposition. Maximum ductility and softness are developed by this treatment.

It is apparent from Fig. 228 that a nickel content of about 8% is sufficient to produce an austenitic * structure in an alloy containing

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* The relationships shown in Figs. 228 and 229 are for commercial alloys containing appreciable contents of carbon and nitrogen. It has recently been shown that pure iron-chromium-nickel alloys are ferritic at high temperatures up to rather high nickel contents.
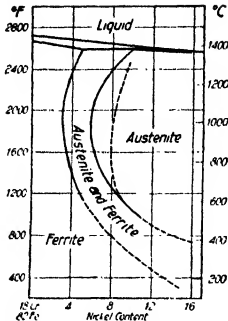
FIG. 229. — INFLUENCE OF NICKEL ON 82% IRON — 18% CHROMIUM ALLOYS. (*Courtesy of V. N. Krivobok and Associates, Trans. Am. Soc. for Steel Treating.*[34])

18% chromium; however, this composition falls very close to the mixed ferrite plus austenite zone. According to Fig. 229 [34] the 18% chromium-8% nickel alloy actually does have a mixed ferrite plus austenite structure at temperatures below about 1200° F. The broken line in the austenitic field of Fig. 229 represents the results of magnetic tests made on samples quenched from elevated temperatures. Magnetic measurements are very useful for studying 18–8 and similar compositions since the development of magnetism in these alloys is proportional to the amount of the magnetic alpha phase present, the gamma phase or austenite being non-magnetic.

The commercial alloys have appreciable carbon contents, hence Fig. 230 [10, 35] is of further interest from the standpoint of carbide solubility. Many features of these diagrams are only approximate and in need of further study and development. Figs. 229 and 230, for example, are not in complete agreement.

For most purposes it is desirable to retain these steels in the austenitic state by rapid cooling from high temperatures within the single phase gamma region. A typical microstructure obtained by quenching from 1950° F. is shown in Fig 231. Coarser grain sizes result upon quenching from higher temperatures (up to about 2200° F. in commercial practice). The homogeneous but twinned austenite grains are comparable to those observed in certain face-centered cubic nonferrous alloys. (See Fig. 71, page 129 for example.)

Ferrite (and carbides) form upon slow cooling from high temperatures or upon cold working after
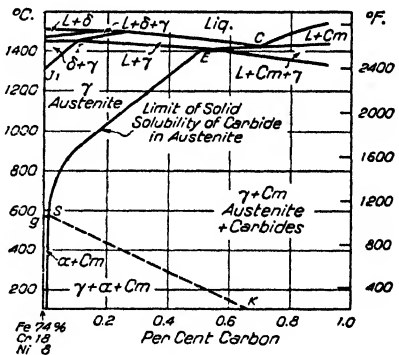
FIG. 230. — INFLUENCE OF CARBON ON STEELS CONTAINING 18% CHROMIUM AND 8% NICKEL. (*Courtesy of V. N. Krivobok, R. A. Lincoln, and R. Patterson, Jr., Trans. Am. Soc. for Metals.*[35] *Originally from The Book of Stainless Steels.*[10])

quenching.   In either case the alloy tends to assume its true equilibrium structure, which consists, according to Fig. 230, of $\gamma + \alpha +$ carbides.   The microstructure of a severely cold worked austenitic alloy is shown in Fig. 232.   Although the ferritic regions are not readily discernible, a marked increase in response to magnetization
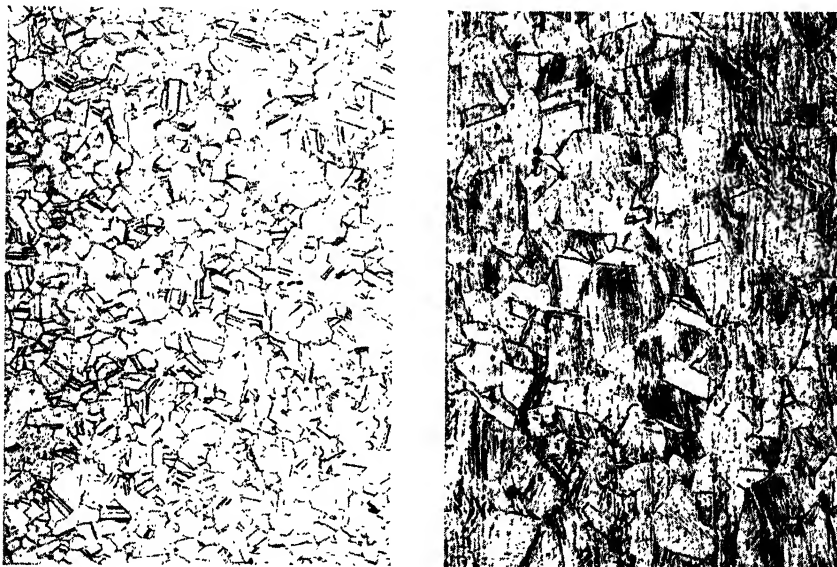
Left.   FIG. 231. — 18% CHROMIUM, 8% NICKEL STAINLESS STEEL QUENCHED FROM 1950° F.   (Etched by electrolysis in 10% perchloric acid.)   Rockwell "B" 80. X100.   (*Courtesy of D. J. Fergus.*)

Right.   FIG. 232. — 18.5% CHROMIUM, 9% NICKEL STAINLESS STEEL QUENCHED FROM 2000° F. AND REDUCED 20% BY COLD ROLLING.   (Etched by electrolysis in 10% perchloric acid.)   Rockwell "C" 30.   X100.

was observed, probably caused by transformation of austenite to ferrite at slip planes.

Mechanical Properties of 18–8. — Tensile strength, elongation, and hardness data for a large number of stainless steel sheets of austenitic compositions have been reported by Krivobok, Lincoln, and Patterson.[35]   Their diagrams for low and high carbon alloys containing 18% chromium are reproduced in Figs. 233 and 234. The high strengths obtainable by cold work are quite evident.   In Fig. 233, for example, a tensile strength of 150,000 lb. per sq. in. with 24% elongation was obtained by 30% cold reduction of an

alloy containing approximately 18% Cr-8% Ni-0.05% C. The increased strength and reduced ductility of the alloy containing 7% nickel is characteristic of the border line compositions which approach or fall slightly within the alpha plus gamma region of Figs. 229 and 230.
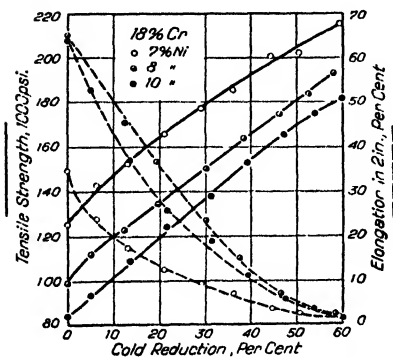
Left. FIG. 233. — SUMMARY OF THE TENSILE STRENGTH AND ELONGATION OF ALLOYS CONTAINING ABOUT 0.05% CARBON, 18% CHROMIUM, AND 7 TO 10% NICKEL, AFTER VARIOUS AMOUNTS OF COLD WORK. (*Courtesy of V. N. Krivobok, R. A. Lincoln, and R. Patterson, Jr., Trans. Am. Soc. for Metals.*[35])
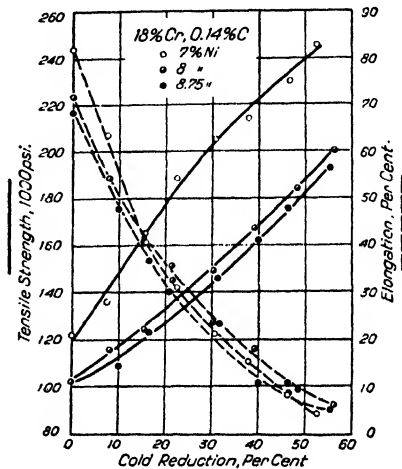
Right. FIG. 234. — SUMMARY OF THE TENSILE STRENGTH AND ELONGATION OF ALLOYS CONTAINING ABOUT 0.14% CARBON, 18% CHROMIUM, AND 7 TO 8.75% NICKEL, AFTER VARIOUS AMOUNTS OF COLD WORK. (*Courtesy of V. N. Krivobok, R. A. Lincoln, and R. Patterson, Jr., Trans. Am. Soc. for Metals.*[35])

The 0.14% carbon steels of Fig. 234 have higher ductility as well as greater capacity for work-hardening compared with the 0.05% carbon steels of Fig. 233. Although higher carbon contents are not associated with improved ductility in ordinary carbon and alloy steels, the effect of carbon on the stability of the austenitic phase is responsible for its unusual behavior in the 18–8 steels. For reasons to be given later, it is often desirable to keep the carbon content as low as possible, in which case nickel contents on the high side (9 to 10%) may be used to increase the stability of the austenite and insure high ductility. When molybdenum and certain other elements are added for specific purposes, the structural condition is

shifted towards the border line state and higher nickel contents are again required to maintain the normal austenitic characteristics.

The high ductility and toughness of the austenitic stainless steels are utilized in deep drawing, spinning, and severe forming operations. Although these steels may be highly deformed without fracture, they work-harden rapidly and may, therefore, require intermediate annealing treatments to facilitate cold working operations. Furthermore, for best corrosion resistance articles which have been
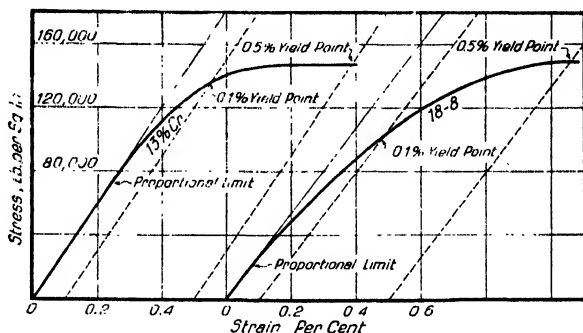
FIG. 235. — TYPICAL STRESS-STRAIN DIAGRAMS FOR HEAT TREATED 13% CHROMIUM, 0.16% CARBON STAINLESS STEEL, AND FOR STRAIN HARDENED 18–8 (0.12% CARBON). (*Courtesy of P. D. Ffield, The Book of Stainless Steels.*[36])

cold worked in fabrication should be annealed. The usual high temperature treatment followed by rapid cooling is used for both purposes.

While the ductility and workability of dead soft austenitic steels is very high, the elastic properties are low, therefore annealed material is not suitable for highly stressed members which must retain their dimensions within close tolerances. The proportional limit, which seems to approach zero stress on annealing, and the various yield strength stresses are very materially increased by cold rolling of sheets or cold drawing of bars and shapes. Tensile stress-strain relations of a typical cold rolled 18–8 are compared with those of a heat treated martensitic stainless steel in Fig. 235.[36] The modulus of elasticity of the 18–8 steel is about 26,900,000 compared with 30,000,000 lb. per sq. in. for the 13% chromium steel. Although these steels have been treated to give the same 0.5% yield strength, 18–8 has a lower 0.1% yield strength (100,000 lb. per sq. in.) and

a very much lower proportional limit.   In the case of cold worked 18–8 these properties, as well as the modulus, may be increased somewhat by a low temperature stress relieving annealing treatment, or by preloading in tension.   Although relatively low elastic properties are typical of cold worked austenitic steel, it does not necessarily follow that the effective strength is low.   Slight perma-
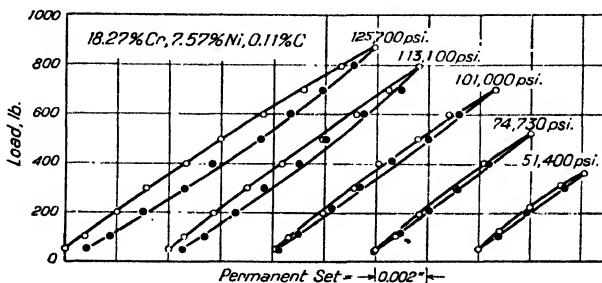
FIG. 236. — STRESS-STRAIN CURVES OBTAINED DURING THE DETERMINATION OF THE PROOF STRESS OF AN ALLOY CONTAINING 18.27% CR, 7.57% NI, 0.11% C.   THE ALLOY HAD BEEN COLD ROLLED 30% AND HAD A TENSILE STRENGTH OF 193,000 LB. PER SQ. IN. WITH 20% ELONGATION.   (*Courtesy of V. N. Krivobok, R. A. Lincoln, and R. Patterson, Jr., Trans. Am. Soc. for Metals.*[35])

nent sets are permissible in most structures and are considered beneficial for reduction of localized stresses.   Yield strength values based on permissible sets of 0.1 to 1.0% are widely used.

Certain specifications (e.g., those of the U. S. Navy) are based on proof stresses, determined by measuring permanent set after release of the testing load.   Test data obtained during the course of a proof stress determination are shown in Fig. 236.[35]   After applying 700 lb. load, corresponding to a unit stress of 101,000 lb. per sq. in., and releasing the load to the small initial setting (50 lb.), a permanent set of 0.0002 inch was observed on a 2 inch gage length.   New samples loaded to higher unit stresses gave permanent sets higher than the specification limit of 0.0001 inch per inch, therefore the proof stress was determined as 101,000 lb. per sq. in.   This testing procedure is comparatively long and difficult and the strain gage used must have high sensitivity and freedom from lost motion on reversal of stress; nevertheless, the uncertainty involved in drawing a modulus line as a basis for proportional limit or yield strength determination is eliminated.

It is apparent that on the basis of yield strength and proof stress specifications cold rolled 18–8 stainless steel qualifies as a high strength constructional material. Furthermore, its corrosion resistance permits the use of light gage sheets in structural members, making it possible for high strength stainless steels to compete with the strong aluminum alloys on a strength-weight basis.

Built-up beams of high rigidity are assembled by spot welding light gage sheets. A welded section used in the construction of railway passenger cars having a high ratio of capacity to weight is illustrated in Fig. 237, a typical floor beam.[37] The success of this type of construction is due largely to the excellent spot welding characteristics of the austenitic stainless steels. Their relatively high electrical resistance and freedom from hardening or embrittlement upon rapid cooling from the welding heat are factors which account for the greater ease of spot welding 18–8 compared with other high strength steels.

FIG. 237.— BEAMS, DRAWN TO SCALE, WITH SAME MOMENT OF INERTIA. THE BOX GIRDER AT LEFT IS WELDED OF 18–8 STRIP, WEIGHS 38% OF 5 IN. I-BEAM OF STRUCTURAL STEEL, WILL CARRY 50% GREATER BENDING MOMENT WHEN FIGURED WITH FACTORS OF SAFETY OF SIX AND FOUR RESPECTIVELY, AND HAS AN ULTIMATE STRENGTH $2\frac{1}{2}$ TIMES AS GREAT. (*Courtesy of E. J. W. Ragsdale, The Book of Stainless Steels.*[37])

Machinability of Austenitic Stainless Steels. — In addition to their tendency to stick to the tool, which is characteristic of stainless steels, the austenitic grades are extremely tough and have a high work-hardening rate. Non-seizing properties are imparted by additions of 0.30% sulphur or 0.25% selenium in the free-machining austenitic steels (Type No. 303). Molybdenum, which is added with the sulphur as molybdenum sulphide, also has a beneficial effect. The extremely high toughness of austenitic steels results in stretching and tearing during chip formation, which causes excessive heating and poor surface finish, a condition which is relieved by phosphorus in the phosphorus-selenium free-machining grade. The

normally high work-hardening rate of 18–8 may be materially re-
duced only by stabilizing the austenite, as, for example, by increasing
the nickel content. Special preparation of cutting tools and reduced
speeds and feeds (compared with carbon steel screw stock practice)
are necessary even for the free-machining grade.[13]

Sensitization of Austenitic Stainless Steels. — The need for
rapid cooling from the heat treating temperature in order to pre-
serve the homogeneous austenitic structure which is stable at high
temperatures has already been stated. According to Fig. 230 the
high temperature quench used in annealing 18–8 is equivalent to the
solution heat treatment of ordinary precipitation hardening alloys.
Unfortunately, however, one cannot utilize a reheating treatment
to harden and strengthen 18–8 because the precipitate, largely chro-
mium carbides, tends to form almost exclusively at grain boundaries
rather than throughout the grains. (This discussion also applies
to 18–12, 25–12, 25–20, and other austenitic compositions.) Fur-
thermore, removal of chromium from the matrix as chromium car-
bides greatly reduces the chromium content of the area immediately
surrounding the precipitated particles, and may cause transforma-
tion of a part of this material from austenite to ferrite.[38] The
deficiency of chromium in the vicinity of the carbide particles makes
the steel sensitive to localized corrosion at the grain boundaries.
According to one theory the localized strain due to precipitation
and phase transformation is partially responsible for the reduced
corrosion resistance.

Carbide precipitation takes place most readily in the temperature
range of 800°–1600° F. upon reheating quenched austenitic steels.
As in other cases of precipitation from supersaturated solid solu-
tions, the lower temperatures require much longer time and give
finer particle sizes. It has also been observed that fine grained
steels require longer time for sensitization than coarse grained
steels.

The normal 18–8 composition is not used for parts operating in
the temperature range mentioned above. In welding practice, how-
ever, conditions leading to sensitization are encountered in the par-
ent metal a short distance from the weld. The deposited metal
itself and the highly heated zone adjacent to the weld are generally

cooled through the precipitation range sufficiently fast (especially in sheets and thin sections) to prevent precipitation, while a zone somewhat farther removed from the weld will be heated to and maintained within the precipitation temperature range for an appreciable time.   Only in spot welding practice is it possible to reduce
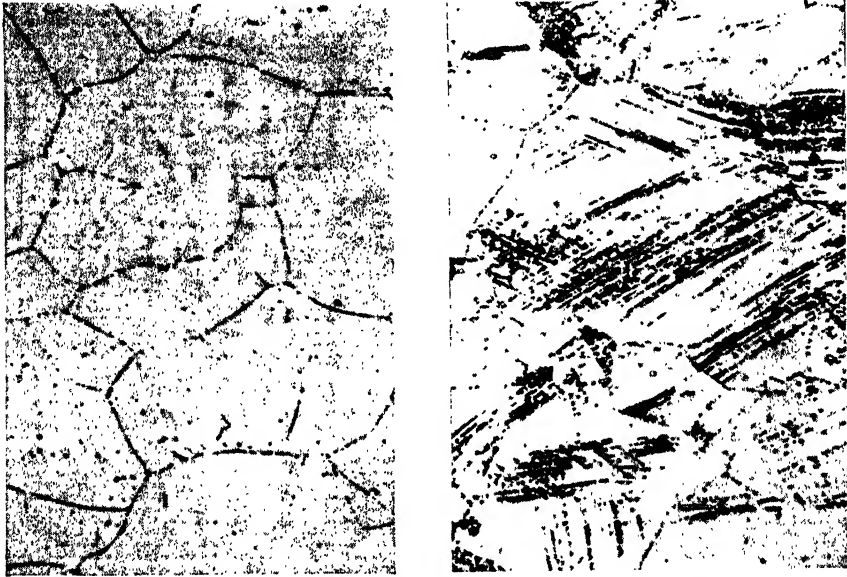
Left.   Fig. 238. — Carbides at the Grain Boundaries of Sensitized 18–8 Steel. (Etched by electrolysis in 10% sodium cyanide.)   X500.   (*Courtesy of D. J. Fergus.*)

Right.   Fig. 239. — Carbides at Slip Planes and Grain Boundaries of 18–8 Steel Heated within the Sensitization Temperature Range after Cold Rolling.   (Etched by electrolysis in 10% sodium cyanide.)   X500.

the time at temperature to such an extent that no harmful precipitation can occur.   If a fusion welded product is to be used under highly corrosive conditions, it should be annealed (quenched from a high temperature) after welding to restore the normal structure, or if this is impractical, a modified composition or a stabilizing heat treatment should be used (to be described in the next section).

The microstructure of a sensitized region in 18–8 steel is shown in Fig. 238.   This condition may vary, under suitable conditions, from precipitation of a small number of isolated carbide particles at the grain boundaries to a continuous network.   Corresponding to

these states the ductility and toughness of the steel decrease progressively while the hardness and strength are appreciably increased.

If sensitization occurs at the surface it may readily be detected by means of special acid treatments which cause rapid deterioration in much the same manner as the test for season cracking of brass. The best known test for stainless steel consists of immersion in a boiling solution of sulphuric acid and copper sulphate known as the Strauss solution.[38] A nitric-hydrofluoric acid solution is also used for test purposes.[39]

After either of these treatments a sensitized steel will tend to lose its metallic ring, or become " dead," and to crack readily upon bending. The electrical resistance increases rapidly in the early stages of intergranular attack, therefore resistivity measurements may be used as a test for sensitization. In extreme cases intergranular attack causes " sugaring " or complete breakdown of the metal into individual angular crystals.

Stabilized Compositions. — Because of the limitations which the possibility of intergranular corrosion placed on stainless steels, especially for welding applications, the efforts of many investigators were applied to the problem of preventing its occurrence. Special stabilized compositions were developed to withstand heating into the sensitization temperature range.

Perhaps the most direct approach to this problem is reduction of the carbon content to values below 0.08%, as in Type Nos. 304, 308, and 309S. Using very low carbon ferrochromium additions, or a special melting practice in which chromium is reduced from chrome ore, carbon contents of 0.03 to 0.08% are obtained. Carbide precipitation diminishes rapidly with reduction in carbon content; however, it is extremely difficult to melt stainless steels with carbon contents low enough to completely eliminate the possibility of intergranular precipitation.

Working on the theory that intergranular corrosion failures were caused not by the carbides present but by depletion of chromium in the vicinity of the carbides, certain investigators were successful in adding elements which have a stronger carbide forming tendency than has chromium, thereby utilizing all the carbon present as special carbides. The success obtained with columbium (Type No.

347) and titanium (Type No. 321) additions has been particularly noteworthy. Since columbium is a strong carbide former and ferrite stabilizer, nickel contents somewhat higher than usual are needed to reduce the proportion of ferrite in the microstructure. It has been found that columbium is not readily oxidized or volatilized during welding (about 75% of the columbium in the rod is retained in the weld [22]), therefore it is effective even in those applications involving " cross welding," or laying of one bead over another, usually at right angles. Titanium, on the other hand, is removed from the deposited layer during fusion welding to such an extent that sensitization may occur adjacent to a second bead as it passes through the first.

Stabilization of austenitic stainless steels may be accomplished not only by modification of the composition but by heat treating procedures corresponding to overaging of other precipitation hardening alloys. By holding for long periods at temperatures within the sensitization range, complete precipitation of carbides and growth of the particles to relatively large size may be effected; furthermore, the time at temperature will be sufficient to eliminate the chromium concentration gradient in the vicinity of the precipitated carbide particles, thus removing the basic cause of chemical attack. Treatments of this nature are best applied after the metal has been cold worked, for in this condition carbide precipitation occurs throughout the grains at slip planes, as well as at grain boundaries, as shown in Fig. 239. This condition hastens the precipitation and growth of carbides, at a given temperature, and reduces concentration gradients. According to Bain's treatment [38], suitable cold working, generally over 15% reduction, is followed by heating to 1250°–1400° F. for an appropriate time. The titanium bearing steel may be heated for 2 to 4 hours at about 1550° F. to stabilize the carbon as titanium carbide. Obviously steels processed in this manner would not have the same high capacity for deformation in fabricating operations as those having a homogeneous austenitic structure. Stabilizing heat treatments have been proposed as a means of restoring immunity to fabricated and welded products without resorting to the high temperature quench required for full annealing.

It should be recalled that normally processed, unmodified austenitic stainless steels are suitable for the majority of applications, and that special steps to avoid sensitization need only be considered for products whose fabrication or service requirements include heating within the precipitation temperature range.

Stress-Corrosion Cracking. — Hoyt and Scheil [39] have recently reported extensive stress-corrosion tests (see also page 186) in which austenitic stainless steels were subjected to highly corrosive conditions while under externally applied stress. The intergranular nature of cracks produced under these conditions was demonstrated and their relationship to prior cold work, carbide sensitization, stabilization by heat treatment, chemical stabilization, and other variables was investigated. It was concluded that only sensitized steels were susceptible to stress-corrosion cracking and that alloys which were stabilized or fully annealed by quenching were not susceptible.

Pit Corrosion. — It is not difficult to understand that localized forms of corrosion can be much more dangerous than general or uniform corrosion. Intergranular attack and pitting are forms of localized corrosion to which numerous ferrous and nonferrous alloys are subject under special conditions of exposure or after improper preparation or treatment of the metal. A pitted surface may be speckled with a large number of superficial depressions or, in extreme cases, perforated with a smaller number of holes of appreciable size.

Pitting is promoted by inhomogeneity in the metal and by irregularities at the surface. The latter may consist of rough finish, scratches, small cracks, embedded abrasives, irregularly deposited corrosion products, or dirt particles. Pitting may also be initiated by concentration of corrosive substances caused by evaporation of drops on the surface of the metal. Pitting often occurs in metals which depend upon protective oxide films for corrosion resistance in liquid media. When this film is broken or destroyed under such conditions that reformation is difficult, due perhaps to the exclusion of oxygen by a foreign substance in contact with the surface, an oxygen concentration cell may be established in which the

exposed zone is anodic while the surrounding oxidized region is cathodic.   Following electrolytic attack of the anodic zone, its depressed condition makes it even more inaccessible to oxygen, thus penetration is progressive rather than self-stopping.   This process

is also known as contact corrosion.   McKay and Worthington [6] discuss the mechanism of pit corrosion and state the conditions under which it has occurred in zinc, copper and its alloys, aluminum alloys, magnesium alloys, nickel and its alloys with copper and iron, iron and steel, and the austenitic stainless steels.

Ferric chloride is one of the most potent reagents for causing pitting in stainless steels.   Brines and other chlorides as well as certain organic substances may also be dangerous.   H. A. Smith [40] has described an accelerated test for pitting in which acidified, 10% ferric chloride is circulated over the test surface under controlled conditions for a period of four hours, after which the number of large pits and the weight loss are determined.   It was found that

FIG. 240. — FERRITE BANDS IN AN AUSTENITE MATRIX IN A STAINLESS STEEL CONTAINING 18.8% CHROMIUM, 8.9% NICKEL, 3.2% MOLYBDENUM, 1.4% MANGANESE, AND 0.04% CARBON.   (Etched by electrolysis in 10% perchloric acid.)   X500.

in most instances this test rated the steels in an order very close to that indicated by service tests.   The steel found to be most resistant to pitting was Type No. 316 (18% Cr-12% Ni-3% Mo), followed by Nos. 302B (18% Cr-8% Ni-2.5% Si), 302, 430, 347, 303, and 321.

The molybdenum bearing compositions have given outstanding service under conditions which cause pitting in the standard 18–8 steel.   In addition to the 18–12-Mo steel mentioned above, compo-

sitions resistant to pitting are also produced by adding molybdenum to 27–4, and other basic compositions.   The 27% Cr-4% Ni-1.5% Mo composition, Type No. 329, is predominately ferritic in structure.

The microstructure of an alloy which may be classified as 19% Cr-9% Ni-3% Mo is shown in Fig. 240.   Its duplex structure consists of bands of ferrite in an austenitic matrix.   Its mechanical properties, as determined on annealed sheets 0.10 inch thick, include 94,300 lb. per sq. in. tensile strength, 50,200 lb. per sq. in. yield strength at 0.5% set, 46.5% elongation in 2 inches, and 85 Rockwell " B."

Several investigators have found that the duplex structures are more resistant than the single phase austenitic structures to both pitting and intergranular corrosion, while their resistance to general attack is probably equal to that of the single phase alloys. Becket [16] reports that chemical composition rather than allotropy controls the corrosion resistance.

The condition of the surface is an important factor in pit corrosion of stainless steels.   It has been found that a buffed and polished surface offers greater resistance to pitting than one which is pickled or finished by grinding after annealing.   Passivation in strong nitric acid or dichromate solution has a similar effect.

Uhlig and Wulff [41] have recently reported the results of extensive investigations into the nature of the surface of stainless steels and its relation to pitting.   They find that the classic oxide film theory is inadequate and offer an electronic theory to explain the passivity of metal surfaces.   Their results indicate that the protective film, if it exists on stainless steels, consists of oxygen mechanically adsorbed at the extreme surface rather than a film of metallic oxides.

Finishes for Stainless Steels. — Finishing operations play an important part in the fabrication of stainless steels.   The cost of preparing the highly polished surfaces required for certain applications of sheets and strip is a major part of the total manufacturing cost. The operations of pickling, grinding, polishing, and buffing are described in detail by Snyder [42] in " The Book of Stainless Steels." General descriptions of the standard finishes for sheets follow :

Finish	Description
No. 1	Hot rolled, annealed, and pickled
No. 2B	Full finish — bright cold rolled
No. 2D	Full finish — dull
No. 4	Standard polish, one or both sides
No. 6	Standard polish, tampico brushed one or both sides
No. 7	High luster polish on one or both sides
No. 8	Mirror finish on one or both sides

Austenitic Heat Resisting Steels. — High chromium contents, as in ferritic Type No. 446, are essential for high resistance to oxidation at elevated temperatures. The principal effect of nickel in heat resisting steels such as 25–12 and 25–20 (Type Nos. 309, 309S, and 310) is improved workability and strength at elevated temperatures, both of which are associated with their austenitic microstructures. Compared with the plain chromium heat resisting steels, the austenitic types are also more satisfactory for welding. On the other hand, the nickel bearing grades are less resistant to the action of combustion products or other gases which are high in sulphur.

According to Lippert [43] the production of 25–12 alloy, Type No. 309, amounted to about 5% of all wrought stainless steel in 1936 and 1937 and 18% of all cast stainless steel (compared with 12% for cast 18–8 and lower percentages of other cast alloys). Most of these heat resisting 25–12 castings are used for furnace parts.

As in the case of the straight chromium or ferritic grades, additions of silicon are useful for reducing oxidation at high temperatures. In addition to Type No. 302B, several compositions having austenitic or duplex microstructures (silicon is a ferrite stabilizer) are manufactured with silicon contents between 2 and 4%.[44] These steels have even higher strength at elevated temperatures than the corresponding chromium-nickel types and are more difficult to hot roll or forge.

Manganese in Stainless Steels. — Between 0.20 and 0.60% manganese is normally present in stainless steels. In free-machining grades the manganese content is somewhat higher (as in carbon steels) in order to offset the bad effect of sulphur on hot working

properties. Up to 2% manganese is used in certain other grades to improve their rolling, forging, and welding characteristics.

Manganese is an austenite stabilizer (see Hadfield's manganese steel, page 380) and may, therefore, be used as a substitute for a part or all of the nickel in the austenitic stainless steels. Several useful compositions have already been commercialized in European countries and further developments are expected because of the possible shortage of nickel in these countries in time of war.

The outright substitution of 8% manganese for nickel in 18–8 gives a duplex rather than a single phase structure upon quenching from high temperatures. The highly ferritic alloys are subject to impact brittleness and reduced ductility compared with the single phase austenitic types. The constitutional diagrams for iron-chromium-manganese alloys indicate that a reduction in the chromium content is necessary to obtain austenitic alloys, even with much higher manganese contents.[16, 45] A 15% Cr-16% Mn low carbon composition, whose structure is essentially austenitic, is more suitable for cold fabrication than the alloys with mixed structures.

The resistance to corrosion and to oxidation at elevated temperatures of chromium-manganese stainless steels is somewhat inferior to that of chromium-nickel steels having comparable microstructures. The austenitic chromium-manganese steels are limited in their high temperature applications because their reduced chromium contents increase scaling losses under oxidizing conditions at temperatures over about 1500° F. However, their strength at high temperatures is said to compare favorably with that of the chromium-nickel austenitic steels and they are more resistant to sulphurous gases.[46, 47]

Useful compositions have been developed in which manganese is used to replace or supplement the nickel in austenitic steels. Copper may also be added for further stabilization of the austenite. Grimshaw[48] has described an alloy containing about 5% manganese and 3% copper added to the basic 18–8 composition. This steel is said to retain its austenitic structure after forging, welding, or even severe cold working and has other distinctive characteristics.

Chromium-Nickel-Iron Alloys. — Alloys with chromium and nickel as the base, rather than iron, are higher in cost than 18–8

and other lower alloy stainless steels but have important applications both as corrosion resisting alloys and as special heat resisting castings for furnace parts and processing equipment.   The requirements of such castings include high structural and surface stability, high load carrying capacity at operating temperatures (creep strength), and resistance to crack-
ing and warping upon repeated heating and cooling.[49, 50]

Approximately 12% of the heat and corrosion resisting alloy castings produced in 1936 and 1937 contained about 35% nickel, 15% chromium (Type No. 330). The high nickel content greatly retards carbide precipitation and other structural changes, making this grade more stable than the lower alloy stainless steels, especially at very high temperatures. Castings of this composition are not recommended for use in sulphurous atmospheres. Stainless steels in which the chromium is in excess of the nickel, such as 30% Cr-15% Ni or the 25–12 grade, may be used, for example, for metallurgical roasting equipment in which the atmosphere is high in sulphur-containing gases.

FIG. 241. — AUSTENITIC CORROSION RESISTING GRAY CAST IRON (NI-RESIST).   (Etched with 3% nital.) X500.

Many other high nickel-chromium austenitic alloys are used in both cast and wrought form for equipment requiring resistance to oxidation, carburization, and nitriding.   These alloys are naturally resistant to corrosion and some grades are also used for electrical resistors.   The most important type is represented by the composition: 60% Ni-15% Cr-25% Fe.[50, 51]   (See also Chapter VI, page 140.)

Corrosion Resisting Cast Irons. — There are many applications for a material having the basic characteristics of cast irons, includ-

ing ease of melting, casting, and machining, combined with greatly improved heat and corrosion resistance. The nickel-copper-chromium iron known as " Ni-Resist " is widely used for such purposes. Its composition range is given as 2.7–3.1% C, 12.0–15.0% Ni, 5.0–7.0% Cu, 1.5–4.0% Cr, 1.2–2.0% Si, 1.0–1.5% Mn, 0.04–0.12% S, and 0.04–0.30% P.[52] The microstructure of a Ni-Resist casting, Fig. 241, consists of flake graphite and carbides in a matrix of austenite.

The tensile strength is about the same as that of ordinary gray cast irons — 20–35,000 lb. per sq. in. Deflections in transverse bar tests are greater than for ordinary irons, the impact toughness is greater, and the hardness is less (120–170 Brinell), all of which reflect the influence of the austenitic matrix on the properties. For special purposes the hardness may be increased to over 200 Brinell by increasing the chromium content to about 6%. Greater hardness, up to about 500 Brinell, may be obtained by chill casting.

Compared with plain gray iron, the austenitic iron is very much superior in resisting growth and oxidation and in strength at elevated temperatures, which accounts for its relatively long life as furnace castings, hotel range tops, manifolds, and other equipment operating at temperatures up to about 1500° F.

The unusually high thermal expansion of austenitic cast iron, together with its wear and corrosion resistance, make it desirable for cylinders and cylinder liners for aircraft and diesel engines using aluminum alloy pistons.

Another special characteristic of this iron is its high electrical resistance, which makes it suitable for cast resistor grids.

In the field of corrosion resisting alloys Ni-Resist is definitely superior to plain gray irons and compares favorably with brasses and bronzes for certain types of service, but does not compete, ordinarily, with stainless steel castings. It is used for pump and compressor castings, pipe, valves, fittings, propellers, and other cast equipment subjected to corrosion in oil refineries, chemical plants, marine service, and numerous other industrial applications.

REFERENCES

1. Marsh, J. S., " The Alloys of Iron and Nickel, Vol. I — Special-Purpose Alloys," The Engineering Foundation, McGraw-Hill Book Company, New York (1938)
2. Merica, P. D., " Constitution of Iron-Nickel Alloys," Metals Handbook, 386, Am. Soc. for Metals, Cleveland (1939)
3. Sands, J. W., " Invar, Elinvar and Related Iron-Nickel Alloys," Metals and Alloys, 3, 131 and 159 (1932)
4. Shubrooks, G. E., " Elinvar Hairsprings for Watches," Metal Progress, 21, 58 (Feb. 1932)
5. Yensen, T. D., " Hipernik and Other Magnetic Alloys," Metal Progress, 21, 28 (June 1932)
6. McKay, R. J., and Worthington, R., " Corrosion Resistance of Metals and Alloys," Reinhold Publishing Corp., New York (1936)
7. Greiner, E. S., Marsh, J. S., and Stoughton, B., " Alloys of Iron and Silicon," The Engineering Foundation, McGraw-Hill Book Company, New York (1933)
8. Kinzel, A. B., and Crafts, W., " The Alloys of Iron and Chromium, Vol. I — Low Chromium Alloys," The Engineering Foundation, McGraw-Hill Book Company, New York (1937) (" Vol. II — High Chromium Alloys " to be published in 1939)
9. Tofaute, W., Sponheuer, A., and Bennek, H., " Umwandlungs-, Härtungs- und Anlassvorgänge in Stählen mit Gehalten bis 1% C und bis 12% Cr," Archiv für das Eisenhüttenwesen, 8, 499 (1934) — and — Tofaute, W., Küttner, C., and Büttinghaus, A., " Das System Eisen-Chrom-Chromkarbid Cr$_7$C$_3$-Zementit," Archiv für das Eisenhüttenwesen, 9, 606 (1936)
10. Thum, E. E. (editor), " The Book of Stainless Steels," Am. Soc. for Metals, Cleveland (1935)
11. Monypenny, J. H. G., " Stainless Iron and Steel," John Wiley and Sons, New York (1931)
12. Parmiter, O. K., " Stainless Steel," Metals Handbook, 533, Am. Soc. for Metals, Cleveland (1939)
13. Palmer, F. R., " Finishing and Fabricating Operations — A, Machining," The Book of Stainless Steels, 99, Am. Soc. for Metals, Cleveland (1935)
14. Krivobok, V. N., " Plain Chromium Irons — Their Toughness and Weldability," Metal Progress, 34, 47 (1938)
15. Franks, R., " Chromium Steels of High Nitrogen Content," Trans. Am. Soc. for Metals, 23, 968 (1935)
16. Becket, F. M., " On the Allotropy of Stainless Steels," Trans. Am. Inst. Mining and Met. Eng., 131, 15 (1938)
17. Jette, E. R., and Foote, F., " The Fe-Cr Alloy System — The Brittle Non-Magnetic Phase," Metals and Alloys, 7, 207 (1936)
18. Krivobok, V. N., " Alloys of Iron and Chromium," Trans. Am. Soc. for Metals, 23, 1 (1935)
19. Monypenny, J. H. G., " European Status of the Stainless Steels," Metal Progress, 29, 68 (June 1936)
20. Comstock, G. F., and Clark, C. L., " Effect of Titanium on Some Properties of 17.5% Chromium Steel," Metals and Alloys, 8, 42 (1937)
21. Becket, F. M., and Franks, R., " Titanium and Columbium in Plain High-chromium Steels," Trans. Am. Inst. Mining and Met. Eng., 113, 126 (1934)

22. Franks, R., "Effects of Columbium and Other Addition Agents on Low Chromium Steels," Trans. Am. Soc. for Metals, **27**, 78 (1939)
23. Andersen, A. G. H., and Jette, E. R., "X-Ray Investigation of the Iron-Chromium-Silicon Phase Diagram," Trans. Am. Soc. for Metals, **24**, 375 (1936)
24. Poboril, F., "Some New Chromium Alloys," Metal Progress, **30**, 56 (Dec. 1936)
25. Palmer, F. R., "The A-B-C of Corrosion and Heat Resisting Steels," Metal Progress, **34**, 340 (1938)
26. Wilten, H. M., "Correlation of Failures from Embrittlement of 4 to 6 Per Cent Chromium Steel with the Notched Bar Impact Test," Trans. Am. Soc. for Metals, **23**, 915 (1935)
27. Wright, E. C., "Five Per Cent Chromium Steels — A. Rolled Material, Especially Tubing," The Book of Stainless Steels, 234, Am. Soc. for Metals, Cleveland (1935)
28. Wright, E. C., "Typical Failures of Still Tubes in Refineries," Metal Progress, **34**, 685 (1938)
29. White, A. E., Clark, C. L., and McCollam, C. H., "Influence of Chromium, Silicon and Aluminum on the Oxidation Resistance of Intermediate Alloy Steels," Trans. Am. Soc. for Metals, **27**, 125 (1939)
30. Newell, H. D., "Properties of the 9% Chromium Steel," Metal Progress, **30**, 51 (Feb. 1936)
31. Bubb, H. D. Jr., "Valve Steels," The Book of Stainless Steels, 485, Am. Soc. for Metals, Cleveland (1935)
32. Krivobok, V. N., and Grossmann, M. A., "Influence of Nickel on the Chromium-Iron-Carbon Constitutional Diagram," Trans. Am. Soc. for Steel Treating, **18**, 808 (1930)
33. Aborn, R. H., and Bain, E. C., "Nature of the Nickel-Chromium Rustless Steels," Trans. Am. Soc. for Steel Treating, **18**, 837 (1930)
34. Krivobok, V. N., and Associates, "Further Studies on Chromium-Nickel-Iron and Related Alloys," Trans. Am. Soc. for Steel Treating, **21**, 22 (1933)
35. Krivobok, V. N., Lincoln, R. A., and Patterson, R. Jr., "Austenitic Stainless Alloys," Trans. Am. Soc. for Metals, **25**, 637 (1937)
36. Ffield, P. D., "Airship Construction," The Book of Stainless Steels, 679, Am. Soc. for Metals, Cleveland (1935)
37. Ragsdale, E. J. W., "Railway Rolling Stock," The Book of Stainless Steels, 694, Am. Soc. for Metals, Cleveland (1935)
38. Bain, E. C., Aborn, R. H., and Rutherford, J. J. B., "Control of Intergranular Corrosion," The Book of Stainless Steels, 388, Am. Soc. for Metals, Cleveland (1935)
39. Hoyt, S. L., Scheil, M. A., "Stress-Corrosion Cracking in Austenitic Stainless Steels," Trans. Am. Soc. for Metals, **27**, 191 (1939)
40. Smith, H. A., "Pit Corrosion of Stainless Steel," Metal Progress, **33**, 596 (1938)
41. Uhlig, H. H., and Wulff, J., "Pit Corrosion of Stainless Steel," Metal Progress, **35**, 247 1939) — and — "The Nature of Passivity in Stainless Steels and Other Alloys, I and II," T. P. No. 1050, Metals Technology, **6**, 1939.
42. Snyder, C. C., "Pickling, Grinding, Polishing, and Buffing," The Book of Stainless Steels, 150, Am. Soc. for Metals, Cleveland (1935)
43. Lippert, T. W., "Stainless," The Iron Age, **141**, 28 (May 5, 1938)
44. Mathews, J. A., "Alloy Modifications of 18–8," The Book of Stainless Steels, 431, Am. Soc. for Metals, Cleveland (1935)

45. Burgess, C. O., and Forgeng, W. D., "Constitution of Iron-chromium-manganese Alloys," Trans. Am. Inst. Mining and Met. Eng., **131**, 277 (1938)

46. Krivobok, V. N., "Permanency of the Stainless Steels," Metal Progress, **34**, 223 (1938)

47. Legat, H. K., "Austenitic Cr-Mn Steels are Unusually Creep Resistant," Metal Progress, **33**, 284 (1938)

48. Grimshaw, L. C., "Manganese and Copper Additions to 18-8 — Armstrong Metal," Metals and Alloys, **6**, 264 (1935)

49. Stanberry, L. J., "Heat-resisting Chromium-nickel-iron Alloys for Furnace Construction," Metals and Alloys, **4**, 127 and 159 (1933)

50. Geiger, G. F., "Higher Chromium-Nickel-Irons," The Book of Stainless Steels, 461, Am. Soc. for Metals, Cleveland (1935)

51. Hunter, M. A., "Electrical Resistors," The Book of Stainless Steels, 502, Am. Soc. for Metals, Cleveland (1935)

52. Vanick, J. S., and Merica, P. D., "Corrosion and Heat Resistant Nickel-Copper-Chromium Cast Iron," Trans. Am. Soc. for Steel Treating, **18**, 923 (1930)

TOOL STEELS

On the basis of mechanical requirements, steels fall rather definitely into three categories. The first group, consisting of low carbon, non-alloyed steels (and irons), is best suited for the many and varied forms of cold fabrication and for welding. While stiffness of the product may warrant special attention, the stress carrying capacity is generally ample for the applications involved. Likewise wear resistance, machinability, and impact toughness are of little moment. In this group are the high tonnage low carbon sheets and wires and a considerable portion of the structural shapes, plates, rods, and bars.

The basic mechanical requirement of the second group is strength; however, toughness, wear resistance, machinability, plasticity, or other special properties may be equally important in a given case. Nearly all machine steels and most structural grades fall in this classification.

A third type of steel is expected to have great wear resistance in rubbing contact with other metals, or the ability to maintain its own shape upon use as a cutting tool or die. Auxiliary properties of tool steels which may demand special attention include toughness, strength, endurance (fatigue strength), machinability (as annealed), and other less easily defined characteristics such as non-seizing properties of dies and freedom from sticking or welding to the chip in the case of cutting tools.

History of Tool Steels. — It is interesting to note that the basic developments in the third or tool steel group generally preceded those in the second and first. The earliest steels were used as tools or hand weapons. They were made from wrought irons by cementation or carburization as mentioned in Chapter VII. The invention by Huntsman in 1740 of the crucible process for melting and casting high carbon steels marked the beginning of the era of modern tool steels.

Alloy tool steels made their appearance in England in 1868 with the invention by Mushet of a tungsten-manganese steel which could be hardened by cooling in air, and had cutting properties much superior to the non-alloyed steels. During the next thirty years improvements in these steels were made and tungsten-chromium compositions were developed having characteristics similar to the original air hardening steels.

In 1898 Taylor and White of Bethlehem Iron Co. (now Bethlehem Steel Co.) discovered that heating certain tungsten-chromium steels to quenching temperatures very much higher than ordinarily used resulted in remarkable improvement in cutting qualities. Modern "high speed" compositions and heat treatments appeared but a few years later as a result of world-wide interest in this discovery, which, more than any other single factor, made possible the quantity production of automobiles and other essentials of the machine age. The history of tool steels has been recorded by Mathews [1], Townsend [2], Bigge [3], and many others.

Carbon Tool Steels. — Despite the widespread use of alloy tool steels and the rising prominence of nonferrous alloys as cutting tools and dies, carbon steels remain the basic material in this field.

Most tool steels are made in electric arc or induction furnaces. All phases of their manufacture are directed toward high quality; thus the cost is greater than that of machine steels, even in the plain carbon grades. However, the following list of warehouse prices of the most common grades (from "Steel," Jan. 2, 1939) shows that the carbon steels (the first three in the list) are considerably cheaper than the highly alloyed grades such as high carbon high chromium die steel and high speed steel. This fact insures consideration of carbon steels whenever tooling costs are an important part of the total cost of production.

Regular	16¢/lb.
Extra	20 "
Special	24 "
Oil hardening	26 "
High carbon high chromium . .	45 "
High speed	69 "

A wide range of characteristics is available in non-alloyed tool steels through variation of the carbon content. For highest toughness with moderate hardness, as required for hammers, rivet sets, heading dies, hot forging dies, and small hand tools, the carbon content may be as low as 0.60%.

Steels with carbon contents sufficient to give eutectoid or near eutectoid structures are used for chisels, punches, and many types of battering tools, shear blades, forging dies, and, to a limited extent, cutters for machine tools.

Carbon steels for use in machine tools as milling cutters, drills, reamers, lathe tools, etc., generally have 1.0% carbon or over. The free cementite of hypereutectoid steels is needed for abrasiveness and maintenance of a keen cutting edge in these applications. As the carbon content increases the toughness diminishes, consequently a proper balance of abrasive hardness and toughness must be sought for each tooling setup. A carbon content of 1.10% is widely used for general purpose carbon steel cutting tools.

Manganese is a most critical element in carbon and low alloy tool steels. As indicated in Fig. 151, page 289, the manganese content for safety in quenching diminishes as the carbon content increases. The values given in this figure are rather conservative since up to 0.30% manganese is present in many water hardening steels with 1.20% C. The 1.42% C steel illustrated in Figs. 242 to 253 contains 0.33% Mn. It did not crack upon water quenching in bar form but might be sensitive to cracking in small or irregular sections. Higher manganese contents are used in oil hardening tool steels, up to 1.75% being present in certain grades.

Silicon is held below about 0.50% in high carbon tool steels because of its tendency to graphitize the cementite. Larger amounts are used in certain lower carbon grades for chisels and punches and in the graphitic die steels.

Phosphorus and sulphur contents are generally very low in high quality tool steels.

Testing of Tool Steels. — It is generally agreed that the most important characteristics of tool steels, their cutting and wearing ability, can only be determined with certainty under actual operating conditions. This state of affairs has long been deplored by purchas-

ers of tool steels who would welcome a less costly procedure involving standardized test values as a basis for selection. This viewpoint was expressed by H. G. Keshian [4] in an A.S.M. paper. In reply J. A. Mathews, now deceased, said in part:

" The tool steel business has always been one in which artistry and skill, coupled with intelligent technical selling, brought success. It is not a mere question of men and machines, and quantity rather than quality. It is the ' divertissement ' of the ferrous industry — pleasure rather than profit. Few industries require so much care, provide such pleasurable variety. offer so much satisfaction or so little hope of great financial reward. The successful tool steel man is imbued with the same spirit of sacrifice as the teacher or missionary. . . .

" I honestly do not know how to start to solve the problem as outlined by Mr. Keshian. For the present, the tool steel industry bids fair to remain one of personal service and one especially responsive to confidence and cooperation which yield mutually satisfactory results to maker and user."

Since this discussion took place (in 1931) notable progress has been made in testing tool steels and the manufacturers have made available more detailed information on the compositions and properties of their products; however, brand names are still widely used and, no doubt, many consumers base their purchases entirely on past experience with certain brands of steels.

Hardness tests are the most generally used of all mechanical tests for checking tool steels in both the annealed and hardened conditions. Tensile tests have little application in this field and are difficult to make on hardened tool steels.

Notched and plain bar impact tests likewise give erratic results which are difficult to interpret. A special torsional impact test developed by Luerssen and Greene [5, 6, 7] is much more satisfactory for materials as brittle as hardened tool steels. Although hardness tests show progressive softening upon tempering hardened tool steels, the torsion impact test is capable of detecting tempering temperature ranges of both maximum and minimum toughness. Palmer [8] finds that the results obtained by the torsion impact machine correlate well with behavior in service. Complete evaluation of this promising test cannot be made until it has been more generally tried and proved.

Fracture tests for grain size and hardenability and deep etching tests for macrostructure, both of which have already been described, are especially applicable to tool steels. The property of tool steels once known only as " body " or " timbre " may now be evaluated within reasonable limits in terms of grain growth characteristics and hardenability as determined, for example, by the Shepherd P–F test described in Chapter VIII. Special hardenability tests used by a prominent tool steel manufacturer are described by Schempp.[9] The influence of initial structure and austenitic grain size on the critical cooling rates of tool steels having widely different hardening characteristics has been reported by Digges and Jordan.[10] The application of hardenability and deep etching tests, as well as the usual mechanical tests, to tool steels is described in " Metals Handbook." [11] One should also read Palmer's book [8] for an interesting and detailed discussion of " timbre " and many other aspects of tool steels.

Heat Treatment. — Tool steels (as well as machine steels) require normalizing and annealing treatments after forging or rolling to put them in proper condition for machining and hardening. As a general rule normalizing temperatures are well above the critical temperature range (above Acm for hypereutectoid steels), while annealing temperatures are only slightly above Ac_1. Detailed recommendations for heat treating the most widely used grades will be found in " Tool Steels " by James P. Gill [12] and in " Metals Handbook." [13]

The hardening of carbon and low alloy tool steels appears to be quite simple. The Ac_1 critical temperature is about 1350° F. and the recommended quenching temperature is generally within the range 1375° to 1550° F., or more specifically 1375° to 1425° F. for plain carbon hypereutectoid compositions. It was formerly necessary to observe the upper temperature limit very closely in all cases to avoid possible growth to coarse grain sizes. This precaution is no longer necessary when controlled fine grained steels having high coarsening temperatures are used, and in the case of alloy grades containing carbide stabilizers it may be advisable to use higher quenching temperatures for better hardenability. Large sections are also quenched from the upper temperature limit.

Selection of a furnace atmosphere for heat treatment of tool

steels is, in general, a more difficult problem than for heat treatment of machine steels. The high carbon content of tool steels increases the tendency towards decarburization and the subsequent development of soft spots upon quenching. In the case of finish machined tools and dies neither excessive scaling nor decarburization can be tolerated.[14] Tools may be packed in charcoal, cast iron chips, or other protective materials to prevent decarburization during heating for quenching. Molten salt baths are also used.

The quenching medium will be determined by the chemical composition, and, to a lesser extent, by the section size of the tool. In the case of cutting tools it is usually desirable to quench for full hardness throughout the section, in which case the critical cooling rate must be surpassed even in the center. Die blocks are not necessarily hardened all the way through; in fact, the toughness of a relatively soft core is often desirable. Internal stresses in quenched tools and dies and methods for their reduction or utilization are discussed by Palmer.[8]

The tempering temperatures for tool steels are low compared with those for machine steels, and are of a much more critical nature. The primary purpose of tempering tool steels is to relieve the brittleness caused by quenching stresses. This may be accomplished with very little decrease in hardness, usually not more than 2 to 5 numbers Rockwell " C."

In addition to the usual precautions to avoid scaling or decarburization, and the need for control of furnace and quenching bath temperatures, heating rates, etc., successful hardening of tools requires the skilled experience of the artisan in preparing the stock and handling it in the furnace and quenching bath. This is particularly true in the case of die blocks of complicated design or irregular section. The combined cost of the steel and its heat treatment may be insignificant compared with the cost of producing such a die; hence cracking, excessive distortion or dimensional changes, decarburization, overheating of thin sections, or other hardening defects may be exceedingly expensive.

Heat Treatment of a Hypereutectoid Steel. — Despite the inherent simplicity of the normal hardening operation there are many interesting possibilities in heat treating a hypereutectoid steel, some

TABLE LXIX. HEAT TREATMENT OF A 1.42% CARBON
TOOL STEEL

Treat-ment No.	Figure No.	Treatment	Rockwell "B"	"C"
A	—	As received (annealed at steel mill)	89	—
B	242	1450° F. — 1 hour — cool in furnace (annealed)	86	—
C	243	1450° F. — 1 hour — cool in air	102	25
D	244	1450° F. — 1 hour — quench in oil	—	35
E	—	1450° F. — 1 hour — quench in oil — 390° F. — 2 hours	—	36
F	—	1450° F. — 1 hour — quench in oil — 570° F. — 2 hours	—	34
G	—	1450° F. — 1 hour — quench in oil — 750° F. — 2 hours	—	35
H	—	1450° F. — 1 hour — quench in oil — 1110° F. — 2 hours	99	20
I	—	1450° F. — 1 hour — quench in oil — 1290° F. — 2 hours	94	—
J	245	1450° F. — 1 hour — quench in water	—	66
K	—	1450° F. — 1 hour — quench in water — 390° F. — 2 hours	—	61
L	—	1450° F. — 1 hour — quench in water — 570° F. — 2 hours	—	55
M	246	1450° F. — 1 hour — quench in water — 750° F. — 2 hours	—	48
N	—	1450° F. — 1 hour — quench in water — 1110°F. — 2 hours	98	19
O	—	1450° F. — 1 hour — quench in water — 1290° F. — 2 hours	93	—
P	247	1450° F. — 1 hour — quench in water — 1290° F. — 20 hours	86	—
Q	—	1600° F. — 30 min. — cool in furnace	88	—
R	248	1800° F. — 30 min. — cool in furnace	95	—
S	249	1800° F. — 30 min. — cool in air	—	30
T	—	1800° F. — 30 min. — quench in water	—	52
U	250	Heated to a white heat — quenched in water	—	54
V	251	Heated to a sweating heat — quenched in water	—	54

All specimens heat treated as $\frac{7}{8}$ inch diameter by 1 inch bars.
Chemical analysis of steel: 1.42% C, 0.33% Mn, 0.30% Si, 0.014% P, 0.014% S.

of which are illustrated by the following photomicrographs and data.

A very high carbon, water hardening, non-alloyed steel was chosen for this purpose. (See chemical analysis in Table LXIX.) As re-

ceived from the mill in the annealed state the hardness was 89 R " B."

All stock to be treated was first annealed at 1450° F. as indicated for specimen B in Table LXIX. The resulting hardness and the microstructure, Fig. 242, were practically the same as before treatment. Unlike the annealed steels of Chapter VII, this microstructure contains but little pearlite and that present is coarse, even considering the high magnification of Fig. 242. The cementite which is ordinarily present in lamellar form in annealed steels appears in spheroidized form in a ferritic matrix. Slower cooling in the furnace would probably have resulted in complete spheroidization. Examination of the equilibrium diagram (Fig. 85, page 161) shows that at 1450° F. the structure consisted of undissolved cementite and austenite. Upon cooling slowly through the Ar_1 critical temperature, the cementite which separated from the austenite apparently contributed to the growth of rounded particles at the nuclei already present, which accounts for the comparative ease of spheroidizing hypereutectoid steels. (See also spheroidizing process (3) on page 203.)

Further heat treatments of annealed bar stock are reported in Table LXIX and Figs. 243 to 251. The hardness tests and microstructures were made after grinding well below the scaled surface.

Air cooling from the hardening temperature, treatment C and Fig. 243, resulted in a pearlitic matrix. At this cooling rate there was insufficient time for spheroidization of the eutectoidal cementite. The excess cementite which was not dissolved at the heat treating temperature remains in spheroidized form. This structure is characterized by greater hardness and poor machinability compared with the annealed steel.

Oil quenching did not harden this steel. The structure shown in Fig. 244 consists of fine (mostly unresolved) pearlite with the usual cementite particles. Reheating to temperatures up to 750° F. produced little change. Softening occurred at 1110° and 1290° F. due to the tendency towards growth and spheroidization of the carbides.

Water quenching resulted in full hardening of the surface of this steel. (The depth hardening characteristics were not determined.) The excess or undissolved carbide particles are present in a marten-

FIG. 242
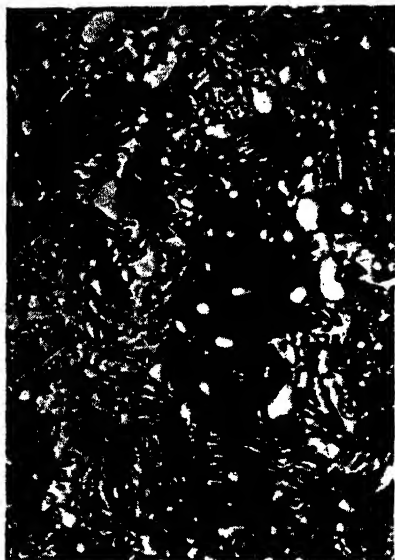
FIG. 243

FIG. 244

FIG. 245

FIGS. 242 to 245. — PHOTOMICROGRAPHS OF A HIGH CARBON WATER HARDENING STEEL. (Etched with picral.) Composition and heat treatments given in Table LXIX. X1500.

FIG. 246

FIG. 247

FIG. 248
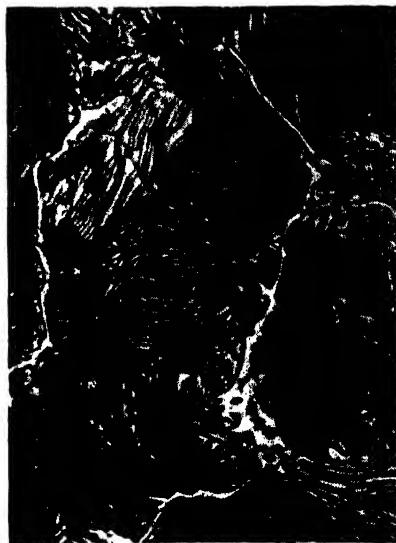
FIG. 249

FIGS. 246 to 249. — PHOTOMICROGRAPHS OF A HIGH CARBON WATER HARDENING STEEL. (Etched with picral.) Composition and heat treatments given in Table LXIX. X1500.

sitic matrix in Fig. 245, which accounts for its extreme hardness of 66 R "C," which is seldom surpassed even by highly alloyed tool steels.

Reheating to various temperatures resulted in gradual reduction of hardness, culminated by the spheroidized structure produced by prolonged heating under the critical range at 1290° F. The structures of the matrices of these steels parallel those of Figs. 113 to 119 quite closely. Only the dark etching sorbitic structure of Fig. 246 and the spheroidized structure of Fig. 247 are shown. It will be recalled that the practical tempering temperatures for most tools are of the order of 300° to 500° F., and at the lower end of this range the microstructure is little changed from the quenched state. The effect of tempering on the structure and properties of tool steel has been studied by Austen and Norris [15] and Engel.[16]

The black areas of Fig. 247 are graphite (not polishing pits) indicating decomposition of cementite during heat treatment. Until recently the presence of graphite or temper carbon in tool steels has been considered to be highly undesirable. Undoubtedly excessive graphitization produces softening and weakening; nevertheless, the beneficial effect of well distributed temper carbon in certain types of die steels has been adequately proved. (Recent data are given by Bullens — Vol. II.[17] See also page 461.)

The remaining treatments of Table LXIX illustrate the effects of heating to temperatures above the normal quenching and annealing range.

Slow cooling from 1600° F., treatment Q, produced a structure similar to Fig. 242 but containing a larger proportion of coarse lamellar pearlite. The presence of spheroidized cementite particles indicated that iron carbide was not totally in solution at 1600° F. Slow cooling from 1800° F., treatment R and Fig. 248, produced a marked change in structure. The excess or proeutectoid cementite became elongated and tended to outline grain boundaries. Part of the cementite was decomposed to temper carbon. The matrix was changed completely to coarse lamellar pearlite. Comparing treatments B, Q, and R it will be noted that the hardness increases as the proportion of lamellar pearlite increases. The effect of various annealing treatments on the structure of hypereutectoid steels has recently been reported by Austen and Fetzer.[18]

Cooling in air from 1800° F. gave a much finer pearlitic matrix and a continuous network of cementite about former austenite grain boundaries as shown in Fig. 249. Quenching in water from this temperature gave a coarse martensitic structure with no excess cementite. The relatively low hardness after this treatment (compared

Left. FIG. 250. — 1.42% CARBON TOOL STEEL, HEATED TO A WHITE HEAT, QUENCHED IN WATER. (Etched with picral.) Rockwell "C" 54. X1500.

Right. FIG. 251. — 1.42% CARBON TOOL STEEL, HEATED TO A SWEATING HEAT, QUENCHED IN WATER. (Etched with picral.) Rockwell "C" 54. X1500.

with treatment J) indicates that untransformed austenite is present along with the martensite.

It is quite evident from the results of treatments R, S, and T that the steel was close to the limit of carbide solubility, Acm, when heated to 1800° F. When carbide networks appear in tool steels, as in Figs. 248 and 249, the toughness is greatly reduced, therefore a treatment should be applied which will restore the excess cementite to a spheroidized condition. The ordinary annealing and hardening treatments will not suffice because the proeutectoid cementite is difficult to alter at temperatures just above Ac_1. Heating to temperatures approaching or above Acm followed by cooling at a rate sufficiently high to prevent re-formation of the network is generally

recommended. (Oil quenching is usually sufficient.) This treatment may be followed by annealing at the normal temperature to restore machinability and refine the austenitic grain size.

Treatments U and V were carried out in a direct fired gas furnace without temperature control. Quenching from an intense white heat gave the structure shown in Fig. 250. The cored dark areas are coarse martensite needles in a matrix which is essentially austenitic.

Heating to a still higher temperature under slightly oxidizing conditions produced oxidation at the grain boundaries or " burning " as shown in Fig. 251. Under more strongly oxidizing conditions grain boundary oxidation may take place at much lower temperatures; in fact, superficial oxidation of grain boundaries has been used as a means of determining austenitic grain size at all temperatures over the critical range. Oxidation or burning seriously impairs the mechanical properties, and if it occurs to any considerable depth there is no method for recovering the steel.

Oddly enough the coarse martensite of Fig. 250 did not develop in Fig. 251. However, it is quite likely that the matrix of Fig. 251 contains untransformed austenite in very appreciable amount.

Alloy Tool Steels. — According to Palmer [8]: " A tool maker should not think of WATER–HARD [carbon tool steel] as ' just another tool steel.' He should think of it as ' home.' Home is the place where a person stays all the time unless he has a good reason for going somewhere else."

The inducements for staying home with carbon tool steel include its low cost, ease of machining, high hardness, and keen cutting edge. There are four principal reasons for going elsewhere and all of these lead the tool maker to alloy steels.

First, despite the high indentation hardness of properly heat treated carbon tool steel, its abrasive hardness as a tool or die may not meet the service requirements. Improvement may be made by increasing the amount and hardness of the excess carbide particles. The alloying elements needed for this purpose are tungsten, chromium, vanadium, and molybdenum.

Second, the toughness of carbon tool steel may be deficient, even though the carbon has been reduced (at a sacrifice of wear resist-

ance). High toughness is a difficult requirement to meet in a steel which is expected to have a long life as a cutting tool or die; however, several alloy steels suitable for battering tools have been formulated by using a proper balance of alloying elements together with a comparatively low carbon content.

Third, machine tools operating at high speeds and taking heavy cuts require a cutting tool which will not lose its " temper " even though it becomes heated by friction to a red heat. The property of retaining hardness upon cooling from such temperatures is known as " red hardness." The possession of high hardness during operation at such temperatures is known as " hot hardness." High speed steels and certain nonferrous alloys such as Stellite and the cemented carbides develop these properties to a high degree.

Fourth, water hardening tool steels present hardening difficulties under many conditions. They are much more sensitive to cracking than oil hardening steels, are shallow hardening, and tend to warp and increase in dimensions upon quenching. In many types of die work it is necessary to finish machine before hardening, hence distortion or even uniform change of dimensions in hardening may make the die worthless. Furthermore, such dies are as a rule most un-uniform in shape and cross section, with holes, sharp edges, screw threads, and projections, all of which increase the possibility of distortion if not cracking in hardening. The so-called " non-deforming " steels withstand heat treatment under these conditions and satisfy the requirements for dimensional accuracy when properly hardened and tempered. In general, they are oil or air quenching steels of high hardenability. They depend upon the retention of the proper proportion of austenite to balance the normal increase in size which accompanies the formation of martensite.

There are several possible combinations of the four special requirements, some of which can be satisfactorily incorporated in a single steel. It is comparatively easy, for example, to combine hot hardness with non-deforming characteristics. In fact, high speed tool steel goes even farther in that it also develops very high wear resistance. On the other hand, high toughness and high wear resistance are definitely opposed.

The extent to which the most widely used grades meet these vari-

TABLE LXX.　CLASSIFICATION OF AMERICAN TOOL STEELS

Courtesy of Am. Soc. for Metals. Metal Progress, Oct. 1938. Adapted from Tool Steels, by James P. Gill

GENERAL PURPOSE CARBON TOOL STEELS

Water hardening; low wear resistance; high warpage; no red hardness; shallow hardening

Carbon	Silicon	Manganese	Sulphur	Phosphorus	Vanadium	Common Uses
0.60–1.25(a)	0.15–0.50	0.10–0.35	0.03 max.	0.03 max.	Note (b)	Almost universal; note (c)

(a) Usually subdivided by 0.10% steps; for instance 0.65–0.75%, 0.75–0 85%, etc.
(b) Plain carbon steels have no vanadium; carbon-vanadium tool steels may have from 0.08 to 0.40% vanadium, depending on grade.
(c) In lower carbon ranges the steels make shear blades, hammers, striking dies, rock drills. In medium ranges of carbon, the steels make chisels, smith's tools, dies, and cutters for machine tools. In the higher ranges of carbon are small cutters, wood workers' tools and cutlery.

CHROMIUM·VANADIUM OR LOW CHROMIUM TOOL STEELS
(Substitutes for carbon tool steels)

Mostly water hardening; low wear resistance; high warpage; no red hardness; medium deep hardening

Carbon	Silicon	Manganese	Chromium	Vanadium	Remarks
0.50–1.40	0.15–0.50	0.10–0.35	0.10–0.25	. . .	Chromium corrects tendency toward soft spots
0.50–1.40	0.15–0.50	0.10–0.35	0.25–0.50		More intense hardness
0.50–1.40	0.15–0.40	0.10–0.35	0.60–1.20	0.10–0.20	Water hardening } Very tough in low carbon ranges
0.50–1.40	0.15–0.40	0.40–0.60	0.60–1.20	0.10–0.20	Oil hardening }

HIGH CARBON, LOW TUNGSTEN TOOL AND DIE STEELS
(Finishing tools for hard steels or nonferrous alloys)

Oil hardening; medium wear resistance; medium toughness; low warpage; no red hardness; medium deep hardening

Carbon	Silicon	Manganese	Chromium	Tungsten	Vanadium (Optional)	Remarks
0.90–1.10	0.20–0.40	0.15–0.30	. . . . .	1.00–1.50	}	Somewhat erratic in heat treatment
1.15–1.25	0.20–0.40	0.15–0.30	. . . . . .	1.75–2.50	0.10–0.25 }	
0.90–1.10	0.20–0.40	0.15–0.30	0.35–0.75	1.50–2.50	0.10–0.25	Water hardening } more dependable
1.15–1.30	0.20–0.40	0.15–0.30	0.35–0.75	2.50–2.50	0.10–0.25	Oil hardening }

MANGANESE OIL HARDENING DIE STEELS ("NON-DEFORMING")
(General purpose tools and especially dies, punches, and broaches)

Oil hardening; low wear resistance; medium toughness; low warpage; no red hardness; medium deep hardening

Carbon	Silicon	Manganese	Chromium	Tungsten	Molybdenum	Vanadium	Remarks
0.85–0.95	0.20–0.40	1.50–1.75	. . . .	. . . .	. . . .	0.10–0.25	More subject to grain growth
0.85–1.00	0.20–0.40	1.15–1.45	0.30–0.60	0.30–0.60	. . . .	0.10–0.25	Corrects above, but hardness is lower
0.85–1.00	0.20–0.40	1.35–1.65	. . . . .	. . . .	0.20–0.35	0.10–0.25	Attains highest hardness
0.90–1.00	0.20–0.40	0.90–1.15	0.50–0.90	. . . . .	. . . . .	. . . . .	Least susceptible to hardening cracks

TUNGSTEN ALLOY CHISEL AND PUNCH STEELS
(Oil hardening steels; shears and battering tools for cold metal; heading dies)

Medium wear resistance; high toughness; low warpage; medium red hardness; medium deep hardening

Carbon	Tungsten	Chromium	Vanadium	Silicon	Remarks
0.45–0.60	0.75–1.25	0.75–1.25	. . . . ..	1.00–1.50	Good wear resistance but somewhat brittle
0.45–0.60	1.50–2.00	0.75–1.25	. . . .	1.00–1.50	Higher tungsten improves wear resistance
0.45–0.60	1.00–1.75	0 50–1.00		. . . .	Low silicon increases toughness 25%
0.40–0.55	1.75–2.25	0.75–1.25	0.10–0.30	. . . . }	Most popular analyses, tough and fine grained
0.55–0.65	1.75–2.25	0.75–1.25	0.10–0.30	. . . . }	

TUNGSTEN FINISHING TOOL STEELS AND DRAWING DIES
(Brittle but intensely hard and keen edges for cutting hard materials)

Water hardening; medium wear resistance; low toughness; high warpage; no red hardness; deep hardening

Carbon	Tungsten	Chromium	Remarks
1.20–1.40	5.00–6 00	. . . .	Slightly better wear resistance than lower tungsten ⎱ High movement; best for drawing dies
1.20–1.40	3.00–5.00	. . . .	Slightly tougher than higher tungsten ⎰ that must be rehardened after wear
1.20–1.40	4.00–6.00	0 40–0.80 }	Chromium improves heat treatability and reduces volume change
1.20–1.40	4.00–6.00	1.00–1.50 }	

SILICON–MANGANESE PUNCH AND CHISEL STEELS
(A water hardening and inexpensive steel for cold cutting)

Medium wear resistance; medium toughness; medium warpage; medium hot hardness; medium deep hardening

Carbon	Silicon	Manganese	Chromium	Molybdenum	Vanadium	Remarks
0.50–0.60	1.80–2.20	0.60–0.90	. . . .	. . . .	. . . .	Spring steel analysis; all high silicon steels liable to soft skin
0.60–0.75	1.70–2.25	0.70–0.90	. . . .	. . . .	. . . .	More carbon gives higher hardness
0.50–0.60	1.75–2.25	0.70–0.90	0.20–0.35	. . . . .	0.15–0.30	Alloys increase hardenability and refine grain
0.50–0.60	1.75–2.25	0.70–0.90	. . . . .	0.40–0.60	. . . .	Molybdenum greatly increases hardenability
0.50–0.60	0.75–1.25	0.35–0.60	0.20–0.40	0.40–0.60	. . . .	Low silicon reduces brittleness and wear resistance

TABLE LXX (*Continued*)

HIGH CARBON, HIGH CHROMIUM PUNCH AND DIE STEELS

(Durable rolls, mandrels, punches, dies, and shears for cold work)

High wear resistance; low toughness; high hot hardness; deep hardening, difficultly machinable

Carbon	Chromium	Vanadium	Molybdenum	Cobalt	Nickel	Remarks
			Oil Hardening Types ("*Non-Deforming*")			
2.25–2.45	12.00–14.00	...	.....	.....		
2.10–2.20	12.00–14.00	0.75–1.00	....	.....		Somewhat tougher
2.15–2.25	12.00–14.00	.	.....		0.50–0.75	Slightly air hardening; most difficult to machine
2.15–2.25	12.00–13.00	0.75–1.00	.....	0.60–1.00	... .	Slightly red hard
			Air Hardening Types (Tougher Than Above, and Deform Less in Hardening)			
1.40–1.50	12.00–13.00	.	0.50–0.60	3.00–4.00	. .	Red hard properties. Good for cutting tools on nonferrous materials
1.50–1.70	16.50–18.00	.	.	.. ..	. .	Liable to harden non-uniformly
1.50–1.60	12.00–13.00	.	0.80–1.00		. .	
1.50–1.60	12.00–13.00	0.80–1.00	0.75–0.90	0.40–0.60	... .	Vanadium imparts greater toughness
1.40–1.55	12.00–13.00	0.80–1.00	0.75–0.90	. . .	0.60–0.80	Most difficult to machine

CHROMIUM DIE STEELS FOR HOT WORK

(Gripper, bending, and heading dies for light work up to 600° F.)

Air or oil hardening; medium wear resistance and toughness; low warpage; medium hot hardness; deep hardening

Carbon	Chromium	Molybdenum	Remarks
0.85–1.00	3.75–4.00	. .	Usually quenched in light air blast
0.85–1.00	3.25–3.75	.	Lower chromium reduces cracks during oil quenching
0.65–0.75	3.75–4.25	.	Oil quenching (lower carbon) but not as rigid at 500° F.
0.85–1.00	3.75–4.25	0.40–0.60	Best air hardener

TUNGSTEN DIE STEELS FOR HOT WORK

(Blanking, forming, extrusion, and casting dies to work up to 1100° F.)

Air or oil hardening; medium wear resistance; medium toughness; low warpage; high red hardness; deep hardening

Carbon	Tungsten	Chromium	Vanadium	Remarks
0.25–0.35	8.00–10.00	2.50–3.50	0.30–0.60	In most general use; serviceable up to 1000° F.
0.35–0.45	8.00–10.00	2.50–3.50	0.30–0.60	Higher carbon gives higher hardness
0.40–0.50	9.00–12.00	1.25–1.75		Chromium lowered to increase toughness
0.25–0.35	12.00–16.00	2.50–3.25	0.30–0.60 }	Increased tungsten raises serviceability to 1100° F.
0.35–0.50	12.00–16.00	2.50–3.25	0.30–0.60 }	Hardenability and brittleness rises with carbon content
0.50–0.60	12.00–16.00	2.50–3.25	0.30–0.60 }	
0.50–0.60	17.00–19.00	3.00–4.50	0.60–1.20	Low carbon, high speed steel

TUNGSTEN-CHROMIUM STEELS FOR HOT WORK AND DIE CASTING DIES

Air or oil hardening; medium wear resistance; good toughness; low warpage; high red hardness; deep hardening

Carbon	Tungsten	Chromium	Vanadium	Molybdenum	Silicon	Remarks
0.40–0.50	6.50–7.50	6.50–7.50	0.20–0.60	....	0.30–0.80	Maximum alloy for hottest services (1000° F.)
0.35–0.45	5.50–6.50	5.00–6.00	. . . .	....	0.30–0.80	Tougher
0.30–0.40	0.75–1.25	4.50–5.00	... .	1.00–1.50	0.80–1.00	Less expensive substitute
0.35–0.40	....	4.50–5.00	... ...	1.00–1.50	0.80–1.00	Properties similar to steel above

HIGH SPEED STEELS

(Cutting tools of all types; tools for severe hot work)

Air or oil hardening; high wear resistance; low toughness; low warpage; high red hardness; deep hardening

Carbon	Tungsten	Chromium	Vanadium	Molybdenum	Cobalt	Remarks
			Conventional Types			
0.55–0.75	17.00–19.00	3.50–4.50	0.75–1.25	....	....	Most used; brittleness and cutting properties vary directly with carbon content
0.55–0.75	19.00–21.00	3.75–4.50	0.75–1.25		....	Better cutting ability but more brittle
0.75–0.85	17.00–19.00	3.50–4.50	1.75–2.25	0.40–0.90	....	Best cutting ability; excellent for finishing cuts
0.55–0.75	13.00–15.00	3.50–4.50	1.75–2.25	....	....	Roughing tools; somewhat erratic in hardening
			Molybdenum High Speed Steel			
0.60–0.85	1.00–2.50	3.50–4.50	0.75–1.25	6.00–8.00	....	Less expensive; "strategic" alloying element used
0.70–0.90	....	3.50–4.50	1.75–2.50	6.00–9.00	....	Improved by high vanadium content
			Cobalt High Speed Steel			
0.65–0.80	17.00–19.00	3.50–4.50	0.75–1.25	....	3.50–5.00	For cutting hard, gritty, or tough materials
0.65–0.80	17.00–19.00	3.50–4.50	1.50–2.25	0.50–1.00	6.00–9.00	Cutting ability varies as total alloy content
0.65–0.80	18.00–21.00	3.50–4.50	1.75–2.25	0.50–1.00	10.00–13.00	Maximum alloy to be forgeable
0.65–0.80	12.00–15.00	3.50–4.50	1.75–2.25	....	5.00–8.00	Good service on special jobs

ous requirements is indicated in a general way in Table LXX. Further information on the properties of these steels will be found in " Tool Steels " by James P. Gill.[12] The balance of this chapter will be devoted mainly to the high alloy grades.

High Carbon High Chromium Steel. — This grade is a comparatively recent development. It will be noted in Table LXX that its general characteristics parallel those of high speed steel rather closely; however, its hot hardness is seldom utilized since its principal applications are as dies rather than cutting tools.

It is necessary to heat to high hardening temperatures in the range 1700° to 1900° F. in order to put a large proportion of the chromium-rich carbides into solution in the austenitic matrix. Upon quenching in oil or cooling in air (depending on the type — see Table LXX) a considerable amount of the austenite is retained, especially at the higher quenching temperatures.[19] The resulting microstructures are quite like those of a quenched high speed steel. (See Fig. 252.) The hardness of this structure is likely to be in the range 61 to 66 R " C " despite the presence of austenite which has not transformed to martensite.

Secondary Hardness. — Upon tempering the quenched high carbon high chromium tool steel at increasing temperatures, the hardness falls off slowly up to about 700° F., whereupon it remains constant or increases slightly up to a temperature of 900° or 1000° F., after which it decreases rapidly.[12] The increase in hardness upon tempering is known as " secondary hardness " or " temper hardness." It is also characteristic of high speed steel and other steels which retain appreciable amounts of austenite upon quenching. It is believed to be the result of decomposition of austenite to a harder martensitic structure which compensates for the natural reduction in hardness due to relief of stresses and breakdown of the martensite originally present. The red hardness of high speed steel and high carbon high chromium steel is attributed to this reaction. The softening which finally occurs at higher tempering temperatures is the result of further decomposition of the matrix to dark etching microstructures similar to the tempered structures of plain carbon steels. The essential difference in the response to tempering of red hard and ordinary steels is the extreme sluggishness of the former.

High Speed Steel. — High speed steel is the principal grade used for cutting metals at high production rates in machine tools. Its importance is reflected in the voluminous technical literature covering its manufacture, heat treatment, properties, and applications. " High Speed Steel " by M. A. Grossmann and E. C. Bain [20] is probably the most complete single treatment of the subject.

The cast structure of high speed steel ingots contains a network of a brittle eutectic constituent, one phase of which is a complex iron-tungsten carbide. This cellular structure must be broken down in the forging and rolling operations so that the carbides finally assume the form of rounded particles well distributed in the matrix. When the carbides retain the cellular arrangement of the cast ingot structure or become segregated as stringers in the direction of working, the toughness and cutting properties are reduced. (Rather low magnifications are best for proper observation of carbide distribution.) This condition is less prevalent in plain carbon and other tool steels because the carbides are more readily dissolved and diffused through the austenitic matrix during hot working.

· The composition of the most used grade of high speed steel is approximately 18% W, 4% Cr, 1% V, and 0.70% C (see Table LXX). The symbol 18–4–1, as applied to this steel, is as well known as 18–8 in the field of stainless steels.

According to Grossmann and Bain [20] the principal carbide present is Fe_4W_2C, thus six atoms are associated with each carbon atom instead of three as in Fe_3C, $(Fe, Mn)_3 C$, etc. The " double carbide " is said to be more stable and harder than ordinary carbide at elevated temperatures. Chromium, vanadium, and molybdenum (when present) are probably associated with tungsten in the double carbides. Cobalt, although not a carbide former, improves the cutting qualities, especially for roughing operations and in cutting hard materials. The cobalt types tend to be more brittle than the 18–4–1 grade.

It is seldom necessary to anneal high speed steels since they are fully annealed at the mill by cooling very slowly from about 1600° to 1700° F. The hardness in this condition is about 225 Brinell (20 R " C "). The structure consists of excess carbides in a dark etching matrix.

Hardening of high speed steel requires heating to even higher temperatures than in the case of stainless steels or the high carbon high chromium die steels. The usual procedure is to heat slowly to 1500° to 1600° F., hold until the temperature is uniform throughout, then transfer to a furnace held at 2200° to 2400° F. The time
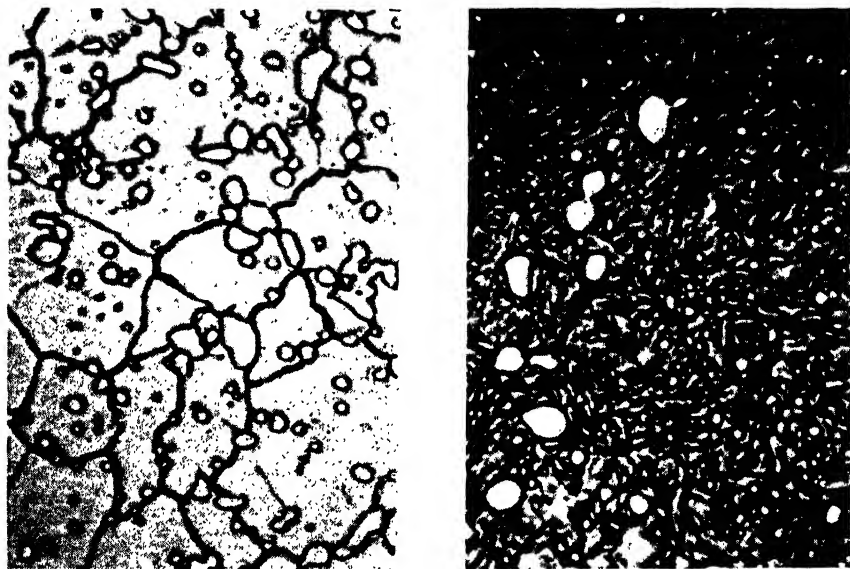
Left. Fig. 252. — High Speed Steel (18–4–1 Type) Quenched in Oil After 3 Minutes at 2350° F. (Etched with 3% nital.) Rockwell "C" 64. X1500.

Right. Fig. 253. — Same as Fig. 252 Reheated for 1 Hour at 1025° F. (Etched with 3% nital.) Rockwell "C" 66. X1500.

at this temperature is short and depends on the size and shape of the tool. A typical treatment for a small lathe tool requires 3 minutes at 2350° F., followed by quenching in warm oil. The microstructure of an 18–4–1 type steel treated in this manner is shown in Fig. 252. From its general appearance one would assume that the matrix is austenitic; however, it is quite magnetic and very hard, 64 R "C." Other evidence also indicates that this structure is essentially martensite with an appreciable proportion of retained austenite.[20, 21] The solubility of the carbides in the matrix, the amount of austenite retained after quenching, and the cutting efficiency increase as the quenching temperature is raised. In general the cobalt

types require higher and the molybdenum types lower quenching temperatures compared with 18–4–1.

Too high a temperature or too long a holding time results in brittleness caused by excessive grain growth and increases the possibility of a burnt or partially melted structure. The grain size of high speed steel is considered by Gill [22], Snyder and Graff [23], and Brophy and Harrington.[24]

It is always advisable to temper high speed steels to improve their toughness. Because of the secondary hardening effect no significant loss in hardness need be sustained at tempering temperatures of 1000° to 1100° F. The hardness of the tempered structure shown in Fig. 253 is 66 R " C " — harder than in the quenched state. This microstructure is more typical of ordinary martensite than the quenched structure of Fig. 252.

As indicated in Table LXX, the molybdenum high speed steels are less expensive (approximately 25% less) and have a definite " strategic " importance in case of a shortage of tungsten. The molybdenum grades have one drawback, a greater tendency towards decarburization in hot working and heat treatment. This may be overcome by using a coating of borax during heating for forging and a properly controlled atmosphere or salt bath for heat treatment.[25] The cutting qualities are comparable to those of the 18–4–1 type for most applications. Further information on these steels is given by Emmons [26], Kennedy [27], and Breeler.[28]

Precipitation Hardening Alloys. — Certain highly alloyed irons are potentially important as tools because they develop great hardness by a precipitation heat treatment. Sykes [29] has investigated this field and finds that the increase in solubility of tungsten in iron with increasing temperature may be utilized as the basis for a solution and precipitation treatment. A typical alloy containing 50% iron, 20% tungsten, and 30% cobalt attains a maximum hardness of 67–69 R " C." Harrington [30] describes an alloy containing Fe, Co, Mo, Cr, V, and C which hardens to about 72 R " C."

Stellite. — Of the nonferrous cutting tools Stellite and the cemented carbides are of greatest commercial interest. Haynes' Stellite, a Co-Cr-W-C alloy, has been used for many years in cast form for cutting tools, surgical instruments, polished mirrors, and as a

hard facing material applied to steel as a welded-on overlay.   It has excellent corrosion resistance as well as abrasive hardness and hot hardness.   Harder and Grove [31] included two Stellite compositions in an investigation of the hot hardness of high speed steels.   Two cylindrical specimens of a given material were placed in parallel position (in line contact), heated to the testing temperature, and loaded while at temperature.   The width of the mutual indentation was used to calculate the hardness.   Compared with the average of results for 18–4–1 steels, an alloy containing 44–46% Co, 32–34% Cr, 14–17% W, and 2–3% C (plus some Fe, Mn, Si, etc.) was somewhat softer up to 1100° F., but at 1292° F. its hardness was 455 Brinell compared with 182 for the high speed steels.

Cemented Carbide Cutting Tools. — Cemented carbide tools are a product of powder metallurgy.   They occupy an important position in the field of cutting tools because they are capable of performance beyond the limits of high speed steels.   They are particularly effective in cutting very hard abrasive materials such as chill cast irons and very tough materials such as Hadfield manganese steel.

According to Sykes [32] the principal steps in the production of these tools are:

"(1) Formation of the metal carbide in the form of powder by heating a mixture of carbon and the metal powder or oxide for several hours at temperature between 1500° and 2400° C., usually in a carburizing atmosphere.

"(2) Intimate mixing of this carbide powder with the binding metal, usually cobalt or nickel, also in the form of powder.

"(3) Compressing this mixture in steel molds of various designs at a unit pressure of from 15 to 30 tons per square inch.

"(4) Initial sintering in some inert atmosphere, such as hydrogen, at a temperature of 800° to 900° C. to impart to the pressed metal sufficient strength for handling and shaping.

"(5) Final sintering at a higher temperature (in the range of 1400° to 1600° C.), which results in a considerable diffusion of the components and a real alloying action.   This operation is also carried out in an inert or carburizing atmosphere."

The tungsten carbide-cobalt type was the first to be developed and is still the most widely used grade.   As the cobalt content increases

from 3 to 13% the strength increases and the hardness decreases. The hardest grades with as little as 3% cobalt are used for fine cuts at high speeds. The 6% cobalt alloy is used for general machining of cast irons, nonferrous alloys, and many abrasive nonmetallic substances. Cobalt contents up to 13% are useful for drawing, extrusion, and shaping dies.

Tungsten carbide tools are not suited to cutting steel because of the tendency for the chips to weld to the tool, ultimately causing crumbling of the edge. Tantalum, titanium, and columbium carbides greatly reduce this tendency. Tools of this type are described by McKenna [3]

Hard alloy carbide particles (including boron carbide) have been incorporated in steel welding rods for use in building up hard surface layers on tools, rolls, dies, cams, and many other wearing parts.[34]

REFERENCES

1. Mathews, J. A., "Modern High Speed Steel," Proc. Am. Soc. for Testing Mats., **19**, Part II, 141 (1919)
2. Townsend, A. S., "Alloy Tool Steels and the Development of High Speed Steel," Trans. Am. Soc. for Steel Treating, **21**, 769 (1933)
3. Bigge, H. C., "Toolsteels, Past and Present," Metal Progress, **32**, 455 and 773 (1937); **33**, 253 and 489 (1938)
4. Keshian, H. G., "Tool Steel from the Consumer's Standpoint," Trans. Am. Soc. for Steel Treating, **20**, 233 (1932)
5. Luerssen, G. V., and Greene, O. V., "The Torsion Impact Test," Proc. Am. Soc. for Testing Mats., **33**, 315 (1933)
6. Luerssen, G. V., and Greene, O. V., "The Torsion Impact Properties of Hardened Carbon Tool Steel," Trans. Am. Soc. for Metals, **22**, 311 (1934) — and — "Interpretation of Torsion Impact Properties of Carbon Tool Steel," Trans. Am. Soc. for Metals, **23**, 861 (1935)
7. Davidenkow, N. N., "Torsion vs. Beam Impact Test," Metal Progress, **30**, 55 (Dec. 1936)
8. Palmer, F. R., "Tool Steel Simplified," The Carpenter Steel Company, Reading, Pa. (1937)
9. Schempp, R., "Quality Tool Steel is Tested for Hardenability," Metal Progress, **30**, 68 (Aug. 1936)
10. Digges, T. G., and Jordan, L., "Hardening Characteristics of One Per Cent Carbon Tool Steels," Trans. Am. Soc. for Metals, **23**, 839 (1935)
11. "Physical Testing of Tool Steels," Metals Handbook, 652, Am. Soc. for Metals, Cleveland (1939)
12. Gill, J. P., "Tool Steels," Am. Soc. for Metals, Cleveland (1934)
13. "Heat Treatment of Tool Steels," (19 articles), Metals Handbook, 987 to 1031, Am. Soc. for Metals, Cleveland (1939)

14. Reiter, F. M., Grenell, L. H., and Adkins, E. C., "Controlled Hardening of Tools," Metal Progress, **32**, 240 (1937)

15. Austen, C. R., and Norris, B. S., "Effect of Tempering Quenched Hypereutectoid Steels on the Physical Properties and Microstructure," Trans. Am. Soc. for Metals, **26**, 788 (1938) — and — Trans. Am. Inst. Mining and Met. Eng., **131**, 349 (1938)

16. Engel, E. H., "The Softening Rate of a Steel When Tempered from Different Initial Structures," Trans. Am. Soc. for Metals, **27**, 1 (1939)

17. Bullens, D. K., and Metallurgical Staff Battelle Memorial Institute, "Steel and Its Heat Treatment," John Wiley and Sons, New York (1939)

18. Austen, C. R., and Fetzer, M. C., "Reactions to Annealing Above the Eutectoid Temperature of Quenched Hypereutectoid Steels," Trans. Am. Soc. for Metals, **27**, 18 (1939)

19. Wills, W. H., "Practical Observations of Some High Carbon High Chromium Tool Steels," Trans. Am. Soc. for Metals, **23**, 469 (1935) — and — **25**, 1013 (1937)

20. Grossmann, M. A., and Bain, E. C., "High Speed Steel," John Wiley and Sons, New York (1931)

21. Emmons, J. V., "Some Physical Properties of High Speed Steel," Trans. Am. Soc. for Steel Treating, **19**, 289 (1932)

22. Gill, J. P., "High Speed Steel, Carbide Segregate and Grain Size," Trans. Am. Soc. for Metals, **24**, 735 (1936)

23. Snyder, R. W., and Graff, H. F., "Study of Grain Size in Hardened High Speed," Metal Progress, **33**, 377 (1938)

24. Brophy, G. R., and Harrington, R. H., "Nature of Abnormal Grain Growth in High Speed Steel," Trans. Am. Soc. for Steel Treating, **19**, 385 (1932)

25. Phillips, A., and Weldon, M. J., "Contributory Effects of Furnace Atmospheres on the Grain-Size of Molybdenum High Speed Steel," Trans. Am. Soc. for Metals, **23**, 886 (1935)

26. Emmons, J. V., "Some Molybdenum High Speed Steels," Trans. Am. Soc. for Steel Treating, **21**, 193 (1933)

27. Kennedy, R. G. Jr., "The Microstructures of Mo-Max and 18–4–1 High Speed Steels Compared," Metals and Alloys, **8**, 289 (1937)

28. Breeler, W. R., "Development in Molybdenum High Speed Cutting Steels," Trans. Am. Soc. for Metals, **27** (1939)

29. Sykes, W. P., "Carbon-free Alloys Hard as Toolsteel," Metal Progress, **32**, 649 (1937)

30. Harrington, R. H., "Precipitation Hardening of Cobalt Steel: A New Tool Material," Trans. Am. Soc. for Metals, **26**, 37 (1938)

31. Harder, O. E., and Grove, H. A., "Hot-hardness of High-speed Steels and Related Alloys," Trans. Am. Inst. Mining and Met. Eng., **105**, 88 (1933)

32. Sykes, W. P., "Cemented Tungsten Carbide Alloys," Trans. Am. Inst. Mining and Met. Eng., **128**, 76 (1938)

33. McKenna, P. M., "Tantalum Carbide Tool Compositions," Trans. Am. Inst. Mining and Met. Eng., **128**, 90 (1938)

34. "Facing or Surfacing by Welding," Metals Handbook, 1172, Am. Soc. for Metals, Cleveland (1939)

APPENDIX

NOTE: John Smeaton (1724–1792) practiced " civil engineery " in England in a
period which offered no special educational facilities in any kind of engineering. Early in
life he abandoned his study of law and reverted to the mechanical bent which had marked
his childhood. His life work was mainly the development and application of water power,
improvement of seaports, and building of a vast system of inland waterways to serve the
growing manufacturing industries. Much of his work on the design and improvement of
mills, pumps, engines, etc., would now be called mechanical engineering.

" To the CARRON COMPANY

" Gentlemen,

" It is from what has occurred to me in the experience of 30 years in the
profession of civil engineery, that I have long entertained an idea that cast
iron anchors, I mean those of the largest sizes, would be found of equal if
not superior strength to those of wrought iron; and though in a case where
not only the lives of men, but the welfare of nations is concerned, no mercan-
tile consideration ought to take place, nor ought any thing to be spared that
can add to the perfection of so very material an utensil as that of the anchor;
yet if, upon a fair and full trial, it shall appear that cast iron ones, of a proper
composition of metal, are in reality equally or preferably to be depended
upon, then the readiness, cheapness, and facility wherewith they are to be
produced, appears to me a very sufficient reason (to say nothing of the en-
couragement of a British·in contradistinction to a foreign production) why
such a fair and full trial should be made as shall be sufficient to put the mat-
ter beyond a doubt.

" I never supposed that any kind of cast iron would be equal in bearing
a stress with that of wrought iron, even of a tolerably quality, provided the
size and shape of the matter to be formed of wrought iron is capable of being
firmly welded, and united in one solid mass; for this reason, I cannot sup-
pose any anchor can be formed of cast iron, that shall bear a stress equal
with one of wrought metal, whose shank is in the smallest part not more than
3 inches, or 3½ inches in diameter: but observing in such large anchors for
first and second rate ships of war, as I have had the opportunity of seeing
when broken, that the wrought bars of which they are composed are very

imperfectly welded and united together in the inside; and having also experienced on the other hand the very great strain that large masses of well-mixed cast iron will bear, when applied to the greatest stresses in mill and engine work, I have been naturally led to put the query, whether beyond some certain medium, that is, whether in those very large and heavy anchors for the largest ships, the substitution of cast iron, instead of wrought, may not be in every respect useful and advantageous.

" Had the trial you have communicated to me, made by the officers of his Majesty's yard at Deptford, appeared to me conclusive, I should have there rested the matter with them, as fully and sufficiently tried; but, with all due deference to those gentlemen, whose knowledge in their profession entitles them to the greatest respect from the public, I beg leave to say, that this is a new case, and therefore till it is tried, in a manner similar to that in which it is to be used, it is in fact no trial at all.

" Had the proposition been to try whether a wrought or a cast iron anchor, or indeed a bar of metal of any size, would best bear the blows of an iron ram or beetle, the mode of trial was perfectly adapted to prove the point; and I am so far satisfied of the fact, as it turned out, that I even wonder that the palms were not broken by the hammers; but I conceive there is nothing like the collision of hard bodies in the real use of an anchor at sea; on the contrary, no stress can possibly be communicated more kindly than that of a ship to its anchor, through the intervention of a long cable. It is possible, in letting go an anchor, it may fall upon a rock, but I conceive an anchor is never let go in foul ground by design, and by choice, but yet it may happen and be necessary. The anchor in its descent having necessarily the cable to haul out after it, and the stock of the anchor, like the log, to haul crossways through the water, the velocity natural to the descent of heavy bodies is hindered from taking place in so great a degree, that let the water be ever so deep, the velocity wherewith it strikes the ground is very moderate, and with this further circumstance that must attend it, that, from the anchor stock and cable both conspiring to act as a rudder, the anchor will necessarily fall with its shank near a perpendicular direction, and therefore have the best chance of impinging upon some part of the crown that is fortified in the best manner to resist blows, as well as every other violence; besides, the rocks below the surface of the sea, being supposed a continuation of the strata above it, and nearly of the same hardness, they are comparative to iron generally soft bodies, and the hardest of them all that lay in masses in this kingdom, that we know of, would, by such a stroke, be bruised, less or more; and this is certain, that the effect of a stroke, where either of the bodies is bruised, or will rebound by elasticity, is widely different from what will happen when neither of them will give way in a sensible degree; I must therefore conclude, that there is not, nor can be, any thing in the real use of an anchor that is in any degree analagous to the stroke of an iron ram, much less to such a stroke applied crossways upon its shank.

" The windmill, axis and oil-press that you cast for me the year before last, the former has withstood the fury of all the storms that have happened since, without the least likelihood of injury; and yet one blow of the Deptford piling-ram, properly directed, would destroy it. The oil-press is in constant work, and every five or six minutes is subject to an alternate pressure and release from it equivalent to 300 tons of dead weight, tending directly to rend it in two; and yet I believe a single well-directed blow of a sledge hammer would break it. If the length of time of the use of these utensils is not thought sufficient, I must add. that in the year 1755, that is 27 years ago, for the first time, I applied them as totally new subjects, and the cry then was, that if the strongest timbers are not able for any great length of time to resist the action of the powers, what must happen from the brittleness of cast iron? It is sufficient to say, that not only those very pieces of cast work are still in work, but that the good effect has in the north of England, where first applied. drawn them into common use, and I never heard of any one failing. Your own method of breaking up the largest iron guns is also an example to the same purpose, where the blow arising from the fall of an iron ball of 7 or 8 cwt. produces an effect that ten times the power of gunpowder would not; for the action of powder, though very quick, yet differs from the instantaneous action of a blow, in much the like manner as a line does from a surface.

" The mode of trial that would appear to me conclusive would be as follows: I would take two anchors, as nearly of a weight and dimensions as possible, the one of wrought, the other of cast iron, not less than three tons weight each (two tons I think too much in favour of the wrought iron for a first trial), and placing them at a competent distance in a right line, with the rings towards each other, for each I would dig a pit in the firm ground, capable of burying both the palms of each anchor: at the bottom of each pit I would fix, edgeways upward, a large elm plank of 10 inches or a foot thick, into the middle of which I would make a moderate perforation of about 3 inches, to receive the point of the fluke of each anchor; these planks I would guard with piles in the securest manner possible, to prevent them, on the application of a great pressure, from moving towards each other, and then well ramming up the whole with earth, so as to bury the anchor, the shanks of each to be inclined upwards, so that the rings may be at or near the surface.

" Then having provided two pair of purchase-blocks capable of purchasing 15 or 20 tons each, with suitable tackle-falls, and cap-stands, crabs, or tooth and wheel gins, I would hook one block of each pair to the ring of one anchor, and the other to the other, securing every thing as much as possible; and the tackle-falls being made to go off to the capstands, each by a snatch block sideways, it is plain, that whatever strain is upon one anchor, the same will be on the other; I would then proceed to heave till something gave way, and which ever of them kept the ground, after the other, by failure, was

drawn out of it, would be the anchor upon which I should be ready, in case of the greatest extremities, to pin my faith.

" The expence of the trial I should think no object, for in case the cast-iron anchor was broke, the other unhurt, there would be no loss but of a little labour, and was I to go to sea, I should chuse the anchor that has been so severely tried, in preference of all others of the same size and kind; but if the wrought iron one gave way, by binding or breaking, so as to quit the ground, while the cast-iron one remained unhurt, then a discovery will result worth the price of 20 anchors.

" In this manner, if the cast-iron anchor proved the conqueror, I would proceed to try those of a lesser weight, so as to find somewhat nearly the medium, at which the wrought iron anchor would have the preference, keeping always on the safe side of the question.

" These, gentlemen, after full consideration, are the result of my genuine sentiments of this subject, which you have desired; and if found useful towards determining a point that may be of much utility to the public service, shall think my pains and study well employed, and remain,

<div align="center">

Gentlemen,
With much esteem and respect,
Your most humble servant,
J. SMEATON."

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Austhorpe, February 7, 1782.

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