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FERROUS METALLURGY

Volume III

METALLOGRAPHY AND HEAT-TREATMENT OF STEEL

THE PENNSYLVANIA STATE COLLEGE

Mineral Industries Series

Van Winkle · MANUFACTURE OF AVIATION GASOLINE

Teichert · FERROUS METALLURGY, 3 VOLUMES

VOL. I. INTRODUCTION TO FERROUS METALLURGY

VOL. II. THE MANUFACTURE AND FABRICATION OF STEEL

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MINERAL INDUSTRIES SERIES

Metallography and Heat-Treatment of Steel

FERROUS METALLURGY

VOLUME III

by

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H. B. NORTHRUP, *Director,*
Mineral Industries Extension.

THE PENNSYLVANIA STATE COLLEGE,
August, 1944.

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INTRODUCTION

Schools of mineral industries are concerned with the exploration, development, and conservation of all mineral resources, and with their production, preparation, processing, and utilization. It is natural that Pennsylvania with its long-established prestige of leadership in the production and processing of mineral products should be the first to organize and operate a school of this type in which all branches of instruction and research are united.

The mineral fuels, the metals, and the nonmetallics are nonreplaceable. Their efficient utilization is therefore of the utmost importance in the industrial life of the nation. The interrelated problems concerned with their production and use constitute a distinct division of education. This may be divided into earth sciences, mineral engineering, mineral technology, and mineral economics. The existence of our mineral industries is dependent upon three divisions of service: resident instruction, research, and extension and correspondence instruction.

Ferrous Metallurgy is presented in a series of three texts, which were prepared originally for extension classes in metallurgy covering a three-year curriculum in that subject. They have been maintained up to date by frequent revisions and improvements and have withstood the test of metallurgical critics and teachers for the past ten years, during which time thousands of students have completed the series.

It is recognized that no one text or series of texts can hope to present fully all existing information on ferrous metallurgy. The aim of this series is to present a curriculum in the subject which shall be fundamental in nature, factual in text, sufficiently extensive in scope, and arranged in a manner that experience has found to be most effective in the classroom and in home-study courses.

The Mineral Industries Series, in which these three texts are included, has been developed as a part of The Pennsylvania State College extension service to meet the current economic and social needs of the nation. There are similar texts in other subjects and advanced, special courses in coal mine mechanization and meteorology. Through this service the principles and truths developed by study and research are translated and carried to the people and applied to the industries. Educational processes taken into the mills, mines, and processing plants result in the promotion of mutual understanding between employers and employees, which is imperative for efficiency, safety, and economy of operation.

EDWARD STEIDLE, *Dean,*
School of Mineral Industries.

THE PENNSYLVANIA STATE COLLEGE,
August, 1944.

FERROUS METALLURGY

VOLUME III

CHAPTER I

INTRODUCTION

In the past 10 years, one of the most important advances in the steel trade has been in the customer's idea concerning the responsibility of the steelmaker for results in fabrication and heat-treatment response.

Before this time the consumer took the stand that if the steel shipped by the mill was of specified analysis, sound, and of good surface, the steelmaker's obligations had been met. But during this decade, in the consumer's plants where heat-treatment was being carried on, continual variations were being found in their results. This condition, therefore, focused attention upon furnaces, quenching equipment, automatic control, and alloy steels, and increased the importance of metallurgical control because of the necessity for reducing the rejections. Even with this attention, and in spite of good analysis, homogeneity, and normal physical properties, there still were found heats of steel that varied considerably in hardening properties.

The automotive industries, along with other consumers of carbon and alloy steels, have now arrived at a point where their technical men require the steelmaker to supply to them metal that will react a certain way when heat-treated in modern, automatically controlled equipment. Further, the reaction of the steel must be uniform from day to day and from heat to heat.

The responsibility rests entirely with the steelmaker and the metallurgist, along with proper supervisory control. In order to obtain the knowledge necessary for a solution of immediate plant problems and problems directly connecting the consumer with the producer, the technical man must make use of the tools and the information developed by the engineer and the scientist in the metallurgical field.

In the light of the previous discussion, this concluding volume, it is hoped, will furnish the reader with a general metallurgical knowledge, so that he may properly evaluate and solve his problems. This involves the determination of the properties of products or raw materials by physical testing, the investigation of any such variances from the desired characteristics, and the correction of these conditions by the use of suitable instruments and technique.

Physical properties having the greatest commercial interest in connection with the metalworking industries are tensile strength, elastic limit, elongation, reduction of area, hardness, impact fatigue, wear and corrosion resistance, and strength at high temperatures. In other words, these studies constitute the evaluation of various results as obtained when a section of iron or steel is subjected to various kinds of stresses and strains.

On casual inspection of a piece of steel, it appears to the ordinary observer to be quite homogeneous, but this same piece of steel upon examination by means of the microscope will be found to be quite complex in structure and possibly not so homogeneous. Such a study is termed a *metallographic* study and the art, the art of *metallography.* A person thus properly trained in the art of metallography, "the study of the internal structure of metals and alloys, and of its relation to their composition, and to their physical and mechanical properties," is invaluable to the iron and steel industry. This art has advanced to such a position that we are now able to discern hidden flaws, in both the raw material and the finished product, without a costly destructive test. Such products as metal castings and welded material can now be studied radiographically, *i.e.*, by the use of the X ray or the gamma ray. Going still further, it has become possible with the use of the X ray to determine the ultimate internal structure or the actual arrangement of atoms in crystals and the distances between atomic centers.

With such modern equipment at our disposal, we are then able to direct our attention to the basic ferrous diagram, that of the iron-carbon system, with reference to cast structures, equilibrium structures, and intelligent application of heat-treatment involving annealing, hardening, and tempering.

With the development in the metal trade, greater and greater demands were being made upon the product, such as lightness and increased strength, combined with other specified properties. These could not be met by the iron-carbon alloys. Such properties could be obtained only by the addition of some element or elements other than carbon or in combination with such other elements and carbon. Such a system is known as an *alloy* iron or steel. In order properly to evaluate these systems, it will be necessary, as in the iron-carbon system, to study the structures and grades along with necessary heat-treatment.

CHAPTER II

MICROSCOPY

SAMPLING

The position from which a specimen to be used for a metallographic study is taken is dependent upon the purpose of the examination. In order to realize the purpose, the specimens taken should be representative of conditions. If a general examination is to be made with no thought of sample preservation, specimens may be taken from end to end as desired, but frequently this is impossible. If the specimen is to be placed in service, the positions available for examination must necessarily be limited. If, for example, a study is to be made of a fracture, proper examination of the fracture itself should be made by taking sectional cuts from the metal in the vicinity of the fracture so that these specimens will offer the greatest possibility of determining the cause of the failure. In addition, a section should be taken at some distance from the fracture, for comparative purposes.

Record cards should be made with such information as the location of the various samples; a stamped number identifying the section or sections taken, which may be cross-sectional, longitudinal, or of some other angular condition; rolling or forging direction; chemical analysis; heat-treatment; or any other data necessary for a complete record.

Specimens can be cut by the use of one of the following methods: a hack saw; a power-driven saw; a cutoff wheel, such as a thin aluminum or carborundum disk; flame cutting in the case of obtaining specimens from large sections; and notching and fracturing in the case of tough and brittle materials. Regardless of the means employed in the sectioning, particular care should be taken to avoid the heating and dragging of the metal at the cut surface. To avoid overheating the specimen is particularly important in the case of hardened steels or alloy steels that are air-hardening, because marked structural differences are brought about by the tempering at the temperatures developed by friction at the surface of the cut. With such steels the difficulty can be minimized by employing cutoff wheels that operate under water. In the case of other materials, this effect can be greatly minimized by the proper lubrication of the saw blade with oil or a cutting solution.

The size of the specimens taken should be kept within reasonable limits (⅜ to 1 in. square or in diameter). This is generally governed by the microscope construction, that is, the space available on the support (stage) for examination; the surface area of the specimen, it being common knowledge that the time expended in polishing a 1-in. square specimen is much greater than the time required to polish say four ½-in. squares; and the thickness, which should be less than any other dimension so that in polishing there will be no tendency to produce rounded edges or extra facets.

Many times specimens are too small in cross section for ease in handling. In such cases, these sections are mounted either individually or in groups in a soft metal, a clamp of suitable design, or in a molding material such as sulfur, Bakelite, Lucite, or Tenite.

The use of Bakelite, Lucite, and Tenite as mounting materials has largely superseded the use of the low-melting-point alloys such as Wood's metal, principally on account of the tendency of such soft alloys to smear in polishing. The plastic materials are particularly suitable, owing to the ease of adapting them to specimens of any shape; to the close contact of specimen and mounting material, which prevents seepage of the etchant with its subsequent staining while drying; to the low molding temperature, which permits its use with even heat-treated materials; to the fact that it is not affected by the etchant; and to the preservation of the flatness of the surface to be examined.

Regarding the other methods of mounting, the following requirements should be noted. The use of steel clamps for mounting in preference to plastic mounting is found desirable when it is essential that the maximum flatness of polished surface from edge to center be preserved. Particular attention must be given, in the use of the clamp, to the hardness of the clamp with respect to the specimen, to the possible interference of the clamp in etching steels of dissimilar composition, and to the possible seepage of the etchant from between the clamp and specimen as the result of the difficulty of obtaining an absolutely tight contact.

The use of sulfur as a mounting material is not desirable because of the ease with which sulfur is dragged over the surface of the specimen, the contamination of the polishing papers and cloth and the possible tempering effect of the specimen in mounting. With certain precautions, however, sulfur can be used with success as a mounting material.

POLISHING

For obtaining a flat surface preliminary to the actual polishing operations, good results are obtained either by employing a file or grinding wheels or papers of the proper grade, or by machining. Grinding wheels, while they can be used when certain precautions are observed, are the poorest practical means for obtaining a good, flat surface because of their tendency to fill up with metal and become dull and grooved in a comparatively short time.

In the grinding operation, regardless of the method employed, the following precautions should be observed: grind the specimen at right angles to the direction in which cutting marks were made, as this is the only positive means of eliminating them; keep the surface absolutely flat by maintaining a light, constant pressure (precaution should be taken to avoid any deep scratches in this stage as certain structures are definitely altered by such working); and avoid any overheating, by keeping the specimen cool with water.

This last precaution is particularly important in grinding cold-worked or heat-treated material. The specimen should never feel warm to the hand, as this signifies quite a rise in temperature of the face being ground, the face temperature being considerably higher than the surface in contact with the hand. The specimen should then be thoroughly washed to eliminate all abrasive material that has collected during the grinding operation.

After a flat surface on the specimen has been obtained, the next step is the polishing operation. This consists in grinding in successive steps with abrasives of increasing fineness until a set of fine scratches is obtained that can be totally eliminated in a short time by polishing with very fine polishing powder. The various methods of attaining this condition may be considered in two steps: first, the rough polish and, second, the finishing polish.

The rough polishing may be carried out in several ways. The first is to use a successive series of abrasives, such as aluminum oxide or magnesium oxide, ranging in fineness from 100 to 600 mesh, suspended in either a water or a water-soap solution. These solutions are used either on a paraffin wheel specially grooved to hold the abrasive or on a paraffin-impregnated canvas or billiard cloth that has been stretched smoothly over the polishing disk while still warm and pliable. The abrasives are applied by sprinkling from a bottle or applying as a moist paste with a brush.

Another widely used method is to grind successively on papers covered with either aluminum oxide (emery) or silicon carbide (carborundum) of increasing fineness. Polishing with papers may be done by hand or on a power-driven wheel. Hand polishing of specimens is considerably slower and more laborious than machine polishing and hence has been supplanted in most laboratories by the latter method. The chief merit of hand polishing lies in the fact that specimens prepared in this manner have a minimum distortion of the surface layer, which is of considerable importance when the specimen is being etched to reveal the true structure of the metal.

Upon the rough-polishing stage depend the length of time required to finish polish, the retention or dislodging of the nonmetallic inclusions, and the thickness of the layer of distorted metal. It becomes quite essential, therefore, that the selection of the type of grinding wheel and of abrasives be carefully made. The final selection to a large extent must be determined by the metallographer. Vilella, in his investigations, finds that finishing with successive papers up to 500 or 600 mesh produces a majority of fine scratches that are easily eliminated in final polishing, but that 2 to 3 per cent are coarse and these tend to lengthen the time. In addition, such dry grinding produces considerable disturbed metal. This method has an important merit, however, in that all nonmetallic inclusions remain intact. Vilella finds that best results are obtained by the use, in the final step, of the paraffin-impregnated billard cloth with 600-mesh abrasive. This method produces fine, uniform scratches; retains the inclusions; and produces a relatively small amount of disturbed metal.

The final polish of the specimen is usually made on a horizontal wheel covered with a suitable material such as woolen broadcloth, silk satin, etc., with a finely levigated alumina, magnesia, or rouge as the polishing medium. The type of cloth and the abrasive used must be determined by the metallographer, as the results obtained with a given procedure will vary considerably from one laboratory to another.

General precautions to be observed:

1. As a general rule the specimen should be rotated and oscillated in such a manner as to eliminate all scratches from the previous operation, along with just enough pressure, in polishing by hand, to keep the specimen flat. In automatic polishing, the pressure is uniformly and constantly exerted by the holder.
2. In every case, specimens should be washed thoroughly after each step in the operation. After the polishing they should be washed and dried thoroughly, care being taken not to touch the polished surface, as a greasy surface will result, which will etch unevenly. Specimens should then be placed in a desiccator or wrapped in cotton, in order that they may be kept free of all moisture until they are ready to be etched and examined.
3. Cleanliness during polishing operations and in the polishing room is of utmost importance.
4. Even with the best metallographic equipment and technique, a poorly prepared sample can produce only a mediocre photomicrograph, lacking in detail and contrast.

Electrolytic Polishing.—This method of preparing iron and steel specimens has been shown to be a very satisfactory substitute for ordinary metallographic polishing.[1,2] Briefly, the procedure is to grind the samples, rub on coarse and successively finer emery to and including No. 000 paper, and then polish electrolytically. The electrolyte employed is made by pouring, a little at a time, 765 cc. of pure acetic anhydride into a cooled vessel containing 185 cc. of 65 per cent perchloric acid (density 1.61) and then adding 50 cc. of distilled water. The mixture should be prepared 24 hr. before it is used.

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[1] JACQUET and ROCQUET, Electrolytic Polishing of Iron and Steel, *The Metallurgist*, June 30, 1939, p. 38.

[2] PELLISSIER, MARKUS, and MEHL, Electrolytic Preparation of Iron and Steel Micro-Specimens, *Metal Progress*, January, 1940, p. 55.

The specimen forms the anode and a plate of either iron or aluminum, approximately twenty times the surface area of the specimen, the cathode. Direct current is used with a density of between 4 and 6 amp. per sq. dm. of anode surface and at least 50 volts as externally applied. During the electrolysis the temperature of the cell must not exceed 30°C. (86°F.). This solution and other conditions noted are satisfactory for all martensitic, pearlitic, and sorbitic carbon steels. The etching time is from 4 to 5 min.

For austenitic steel, the same solution may be used with a current density of at least 10 amp. per sq. dm. It is recommended, however, that a solution of acetic anhydride and perchloric acid (65 per cent), in the ratio 2:1, be used with a current density of 6 amp. and 50 volts. The etching time will be from 4 to 5 min.

It is important to note that mounting is unnecessary and that mounting materials of organic or synthetic nature such as Lucite, Tenite, and Bakelite should not be used because of the explosive reaction between these materials and perchloric acid.

Surfaces prepared in this manner after etching with the usual reagents are entirely satisfactory for microscopic examination and photography at all magnifications. The method not only is rapid but, comparatively speaking, requires very little skill. The specimens are more readily polished to an edge than with ordinary polishing procedures, and specimens of any form and dimension may be polished on all their faces. This method seems to have a special advantage for the study of nonmetallic inclusions, since they suffer less damage than in mechanical polishing and, hence, are more clearly defined.

MICROETCHING

Subsequent examination of a well-polished surface, in general, reveals no trace of its structural characteristics, either visual (macro) or optical (micro). In order to afford a proper insight into the structure, the polished surface must be treated in such a manner as to "render visible the chemically or physically distinct parts and to render more or less increased optical contrast between the various structural components."[1]

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[1] Rosenhain and Haughton, "Introduction to Physical Metallurgy," 3d ed., Constable & Company, Ltd., London, 1935.

The usual method for the development of the metal's crystalline structure for microscopic examination is to employ a chemical reagent.

If a properly prepared sample is examined microscopically prior to the application of the etchant there will be few, if any, structural details apparent. Such an examination, however, may be of value since surface defects such as nonmetallic particles that have not been smoothed over during the polishing can be studied. The absence of a crystalline structure is believed to be due to an extremely thin, opaque, amorphous, metallic film covering the polished surface. This film apparently is the result of the severe mechanical strain and cold work during the polishing, which breaks up the crystalline structure of the topmost layer of metal and converts it into a hard layer of more or less amorphous metal. This effect is particularly important in those steels that have a high work-hardening rate, such as stainless steel.

In subsequent etching with a chemical reagent the first action must be the complete removal of the amorphous layer. After its removal the etchant may then slowly attack and bring about the development of the true structure of the metal. The action of the etchant is dependent upon the chemical and structural nature of the specimen, the chemical nature of the reagent, the time of etching, the temperature of the reagent and the specimen, and the agitation of the reagent during the process.

Reagents in all types of etching act in the same way; that is, they slowly attack and dissolve the metal. The manner in which the reagent attacks is dependent upon the material itself, whether it consists of a single constituent in which all parts of the metal are of exactly the same chemical composition or whether it is complex in nature, where there are several different constituents each differing in chemical composition from the others.

In the case where a pure metal is being etched, the chemical attack of the reagent does not take place uniformly over the exposed grain surface but is found to attack each grain selectively along certain crystallographic planes. As a result the rate of attack will vary according to the orientation of the grains with the surface and a grain contrast will be evident. This contrast is due to the fact that the planes that are parallel in the same grain are not necessarily parallel to those in neighboring grains.

In the case where an alloy composed of two or more structural constituents is being etched, the reagent is selected so as to be selective in its attack. The result is usually partial attack and roughening of one constituent while the other remains unaffected. This allows certain of the constituents to stand out in relief and produces, when observed under the microscope, what might be termed a *shadow* effect. The unaffected part will be bright in color while the roughened, depressed part will appear dark.

The reason for the appearance of the structures when examined microscopically after etching is illustrated in Fig. 1-II. In *A* the polished viscous layer appears bright with no detail, owing to the fact that the light rays from the vertical illuminator of the microscope strike the surface normally, with the result that they retrace their path. After proper etching, the grain contrast becomes apparent as shown in section and plan in *B*. In this case one grain reflects most of the light up the microscope tube, while an adjacent grain reflects most of the light in other directions and will appear dark. When the structure is duplex in character, as for example pearlite in steels, the differential attack of the etchant will bring about a condition as in *C*. The cause of the apparent difference in size of the duplex constituents is also shown in *C*. The grain *X* appears to have wide laminations because of the angle at which the surface was cut, when actually the true spacing is still the same as in the adjacent grain.

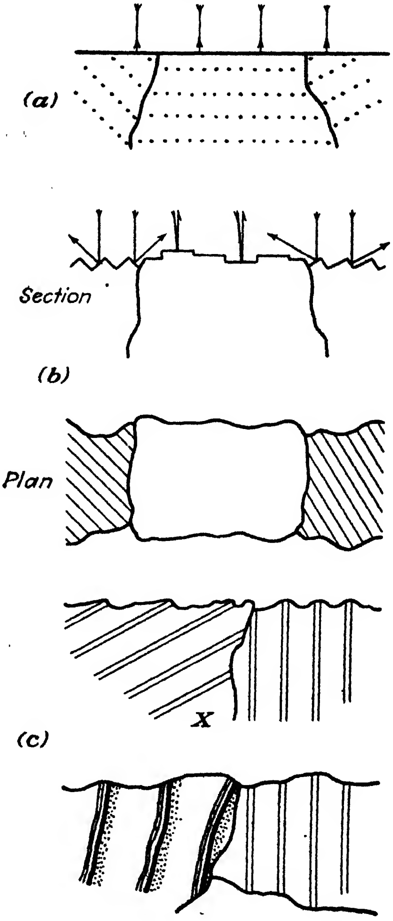

Fig. 1-II.

Etching Reagents.—Etching reagents are basically dilute solutions of organic or inorganic acids, alkalies, or other solutions complex in nature. As has previously been noted, the final selection of a solution to bring about the structural development desired is dependent upon the composition and structural condition of the metal or the alloy.

For the development of the microstructure of iron and steel, solutions of nitric acid and picric acid are the ones most widely used and will be the only ones herein discussed. For a detailed study of etchants for microscopic examination of iron and steel and alloy steels, reference should be made to the 1939 edition of the American Society for Metals "Metals Handbook."

Nitric Acid and Alcohol (nital).—The composition of this etchant for general application varies from 1 to 5 per cent concentrated nitric acid (sp. gr. 1.42) with ethyl or methyl alcohol. A 2 per cent nital solution may be prepared by mixing 2 cc. of the acid with 98 cc. of alcohol. The concentration of the solution is varied according to the desired results. The etching rate will increase with increasing acid concentration, while the selectivity increases.

Picric Acid and Alcohol (picral).—In general the composition most widely used is the 4 per cent solution. This is prepared by mixing 4 g. of clean, dry picric-acid crystals with 100 cc. of methyl or ethyl alcohol.

The use of nital and picral may be summarized as follows:[1]

Use 4 per cent picral

1. To show maximum details in

a. Pearlite	*c.* Tempered martensite
b. Martensite	*d.* Bainite

2. To detect undissolved carbides in martensite.
3. To differentiate between ferrite, martensite, and massive carbide by coloration.
4. To differentiate bainite from fine pearlite.
5. To detect carbide particles in the grain boundaries in low carbon steel and wrought iron.

Use nital—1 to 5 per cent solutions

1. To develop ferrite grain boundaries in low carbon steels.

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[1] VILELLA, J. R., "Metallographic Technique for Steels," A.S.M., Cleveland, Ohio, 1938.

2. To produce maximum contrast between pearlite and a cementite or ferrite network.

3. For chromium bearing, low alloy steels, resistant to the action of picral.

4. To develop boundaries of ferrite in structures made up of martensite and ferrite.

Etching Procedure.—The procedure and manipulation involved in the etching process requires a certain amount of skill, which is relatively easy to develop. Certain factors other than the selection of the proper reagent must be recognized to produce satisfactory results. These factors are the method of applying the reagent to the surface, the approximate time of reagent contact with the surface, and reagent temperature.

In general, the application of the etching reagent is made either by immersion or by swabbing. When immersion in the solution is used, the specimen is held in the solution by a pair of tongs for the required length of time. Constant movement of the specimen is recommended so that air bubbles clinging to the surface may be eliminated and so that a film of fresh reagent will be in contact with the etching surface at all times.

If the specimen is etched by swabbing, the polished surface is swabbed rapidly for the required period of time with a tuft of cotton that has been saturated with the etching solution.

After the elapse of the required etching time, the specimen is thoroughly washed under a jet of water in order to remove the reagent and arrest the attack. This is quite essential, not only in arresting the action, but in preventing the formation of acid salts that would disfigure the surface. A thorough cleaning under a jet of hot water, with its evaporation and removal by violent shaking or careful wiping with a soft cloth, will suffice. Where greater care is necessary, the specimen should be washed in absolute alcohol to remove the water, and the alcohol should then be evaporated in an air stream with a fan. If the etching time has been insufficient, the best procedure is to repolish until all traces of the etch have been removed, and then to reetch.

The surface as properly etched and dried will remain useful for several days or even weeks if it is kept in a desiccator provided with the proper drying agent (calcium chloride).

Time of Etching.—The first important factor in correctly etching a specimen is the selection of the reagent best suited to the chemical composition and the physical condition of the metal, while the second is suitably controlling the time of etching to produce the proper degree of contrast between the different components of the structure. The time element is dependent upon the structure to be emphasized and upon the magnification at which it is to be photographed.

Depending upon the etching reagent selected, the etching time will vary from a few seconds to a number of minutes or even longer. It is difficult to give precise instructions covering all cases, so the following recommendations are made.[1]

1. Etch no deeper than is necessary to bring out the significant detail.
2. A degree of contrast satisfactory for low-power photomicrography is generally excessive for the best possible definition of the structure at high magnification.
3. If high contrast is desired, it is preferable to obtain it by photographic means rather than by deep etching, for deep etching obliterates the fine details of the structure.
4. If a specimen is insufficiently etched (too-low contrast) it is preferable to repolish on the finishing wheel and to reetch rather than to superimpose one etch on the other.
5. Do not touch an etched surface with anything, and photograph soon after etching.

Electrolytic Etching.—Several different reagents may be used for electrolytic etching, each of which has its advantages in developing specific structures. After suitable polishing, the specimen is placed in the chosen electrolyte and with the specimen as the anode and either a stainless or a platinum plate as the cathode held ¾ to 1 in. apart, a current is supplied by either four dry cells in series or an equivalent 6-volt source. The etching time with each reagent is dependent upon the structure to be developed.

The following compositions and applications are the most widely used.

Chromic Acid.[2]—The electrolyte is prepared by dissolving 10 g. of CrO_3 in 100 ml. of distilled water. It is used for developing

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[1] Vilella, J. R., Photomicrography, A.S.M. "Metals Handbook," p. 176, Cleveland, Ohio, 1939.

[2] Baeyertz, M., The Use of Chromic Acid in Electrolytic Etching of Iron and Steel, *Trans. A.S.M.*, **25** (No. 4), 1185 (1937).

various structures, except grain boundaries of ferrite. It attacks cementite and chromium carbides of stainless steels very rapidly, austenite less rapidly, and ferrite very slowly if at all.

Oxalic Acid.[1]—This electrolyte is prepared by dissolving 10 g. of oxalic acid in 100 ml. of distilled water. It is particularly adaptable in revealing both carbides and grain boundaries of the austenitic stainless steels and high nickel alloys.

Sodium Cyanide.[2]—This electrolyte is prepared by dissolving 10 g. of sodium cyanide in 100 ml of distilled water. It is particularly adaptable for detecting the presence of precipitated carbides (chromium carbides, tungsten carbides, etc.), and will darken these without attacking the austenite or grain boundaries.

MACROETCHING

The development and application of the macroetch, which consists in subjecting the metal to the action of an acid to bring out the physical and chemical nature of the material for visual examination, has been rapid and has attained a distinct place in industrial routines. Its popularity is due to its relative simplicity, to the simple equipment necessary, to the easily acquired technique, and to the quickness and readiness with which it reveals many serious internal defects.

In performing any type of etch test there are two important procedures that must be followed: the carrying out of such a test under standard conditions, and the proper classification of the structures revealed.

Standard Conditions for Macroetching. *Preparing the Sample.* In the preparation of the sample, the size, shape, and surface are of importance. Large sections are generally difficult to handle in acid solutions, and sections that are extremely long or narrow require excess acid and odd-shaped containers. Usually thin disks or slabs are found to be most convenient.

In general, in sectioning a piece of metal for inspection, proper consideration must be given to the type of defect, that is, internal or external. For external surface, little or no previous prepa-

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[1] ELLINGER, G. A., Oxalic Acid as an Electrolytic Etching Reagent for Stainless Steels, *Trans. A.S.M.*, **24** (No. 1), 26 (1936).

[2] ARNESS, W. B., The Sodium Cyanide Metallographic Etch Test for Revealing Precipitated Carbides in the 18-8 Type Stainless Steels, *Trans. A.S.M.*, **24** (No. 3), 701 (1936).

ration is necessary as the scale on the surface will be removed by the acid and render the surface defects visible. For internal surface, a section must be cut through the specimen at the point where inspection is desired.

As for the surface on which the pattern is to be revealed, it is essential to have it of such a finish that, following the action of the acid, there will be little or no trace of toolmarks that would mask the details and make interpretation of the results impossible.

Polished surfaces are not essential, but the surface produced by a hack saw, even a fine one, is not desirable as the cut marks often appear on the etched surface. A recommended type of surface may be had by filing the cut surface with a fine file and possibly finishing with No. 0 or 00 emery paper. Very satisfactory results, however, have been obtained by using surfaces prepared by grinding on a soft or a medium-grade grinding wheel. Care must be taken to avoid burning.

Acid Reagents.—The acids with which we are chiefly concerned are hydrochloric, sulfuric, nitric, and picric. These may be used in a wide variety of concentrations in alcohol or water, and together in a large number of different mixtures. For detailed information on the composition of the various solutions, their intended purpose, procedure, and uses, it is recommended that reference be made to the compilation shown in the A.S.M. "Metals Handbook."[1]

The solution most widely used in ferrous work is a 50 per cent hydrochloric acid in water. This solution undoubtedly has greater selective etching properties than any other single acid and its concentration remains stable on heating. The fumes from this mixture are quite corrosive so that the work had best be done under a hood. It is also most advisable to use rubber gloves when handling specimens for all macroetching.

The effect, as is true of that produced by all macroetching reagents, is dependent upon the solution and strength of the reagent and the time and temperature of etching. These factors, as will be found through experience, are variable for the same material. According to Rawdon, it is a general rule that the strong acids, which etch in a short time, give better contrasts than dilute acids which need a longer time.

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[1] A.S.M. "Metals Handbook," pp. 726–735, Cleveland, Ohio, 1939.

Sulfur distribution is usually shown by the sulfur-print method. A mat-finish photographic paper is moistened with a solution of 2 to 3 per cent H_2SO_4 (concentrated) in water. The specimen, whose surface has previously been smoothed, is placed face down on the paper and pressed firmly so as to assure uniform contact. Each sulfide inclusion reacts with the acid to evolve sulfuretted hydrogen (H_2S) which in turn reacts with metallic salts in the paper, in this case silver, causing a brown mark (silver sulfide), which indicates the sulfur-rich areas. The paper is then washed and fixed in a thiosulfate solution.

Etching Time.—Etching time, the time allowed for proper action of the acid reagent, is a variable quantity. This is due to the number of varying factors, such as type of steel, whether rimmed or killed; chemical composition; heat-treatment; etc. The final point in the action must depend upon the completeness of the details or characteristics inherent in the steel.

Etching Temperature.—Etching temperature, the temperature of the reagent, is as important as etching time. In order to avoid troublesome exactitude in temperature measurement, a practical standard has been selected—the temperature of boiling of the reagent. Thus control of the bath condition necessitates only the observation and regulation of the boiling. It is recommended first to clean the specimen and then to heat it in boiling water to the same temperature as the acid. This results in better control of etching time and also aids in duplicating conditions.

Following the development of the structure, the specimen is removed from the solution and washed thoroughly under running water, the dirt and slimy carbon deposit being removed by scrubbing with a soft brush. After thorough rinsing, the surface should be dried by playing warm dry steam or air on it or by placing it on a hot plate for a few minutes, avoiding the effects of unnecessary heating and corrosive effect of the gas. This procedure will ensure a good surface condition, which will not become tarnished for many days. For longer preservation, the surface may be covered with a thin coat of transparent lacquer or a film of oil. Tarnished specimens may be rejuvenated by immersion, soaking, and rubbing with 5 to 20 per cent sodium or ammonium citrate solution, following by water washing and drying.

Interpretation.—In order to interpret correctly defects observed on macroetched sections, it is necessary to have knowledge, experience, and good judgment in regard to discontinuities and nonhomogeneity in the metal and to the chemical and mechanical technology of iron and steel. Every piece of steel is a heterogeneous mass composed of mixtures and chemical compounds and mixtures of both metallic and nonmetallic segregates. All steels are subject, therefore, to a selective attack by inorganic acids, different acids attacking different compounds or mixtures. When etched with a given acid, the steel may appear solid and free of inclusions, while with another it may be spongy and contain many inclusions. Hardened steel will etch differently from annealed steel, and alloyed steels differently from those that are unalloyed.

The deep-acid test reveals readily and quickly many of the characteristics of the steel, but in some instances the characteristics require further examination by standard physical testing methods and the microscope to evaluate them. The knowledge that a piece of steel is nonhomogeneous is in itself insufficient evidence on which to base a definite conclusion as to the properties of the material. The real value of the test lies in separating materials obviously defective from those that are not, and further, in indicating characteristics that are not detrimental to the material but that give an indication that further investigation is necessary.

Characteristic Structures and Defects. *Cracks from Etching.*—Highly stressed steel, *i.e.*, unannealed, rolled, or forged tool steel and hardened tool steel, is subjected to cracking by the etching solution itself.

Open Grain.—Open grain may be defined as surface roughness or possibly sponginess. This appears to be due to the method of deoxidation that controls the composition, size, and distribution of the oxide inclusions. Because of this, steels high in oxide content often exhibit, when subjected to the deep-acid etch, less "open grain" than do clean steels. When this characteristic is in evidence, it is advisable to examine the steel at low magnification to determine accurately if there is an abnormal amount of nonmetallic sulfide or silicate present.

Pipes and Bursts.—Pipes are internal cavities and are invariably associated with segregated impurities, which are deeply attacked by the reagent. Material that has burst as a result of improper forging, when it is subjected to the test should not show sponginess around the burst, while material piped will exhibit sponginess in the area adjacent to the pipe.

Segregations—Nonmetallic and Metallic Segregates.—Cross sections of bars and billets subjected to the test often show varying degrees of porosity, which in some cases becomes so extreme as to cause the observer to classify the material as piped. In general, porosity is the result of selective etching of both metallic and nonmetallic inclusions or segregates.

In the unkilled steels—those not fully deoxidized—numerous blowholes and nonmetallic inclusions will be shown distributed throughout the section. Blowholes and nonmetallic inclusions will appear as pits. The presence or absence of certain nonmetallic inclusions may be determined by the use of several different reagents or combinations of reagents having selective attacks.

In killed steels, nonmetallic inclusions usually appear as pits and should not be confused with pits occurring from the etching of metallic segregates. When nonmetallic inclusions are indicated, as for example in high alloy steels, which also would contain metallic segregates or alloy carbides, then the steel sample should be hardened and etched for comparison. If the pits have been caused by a metallic segregate in the annealed sample, hardening dissolves a sufficient amount of the segregate or carbide present, so that the hardened sample will appear different after etching. Etched steels containing nonmetallics will appear much the same both in the hardened and in the annealed condition.

Internal Cracks.—Internal cracks, sometimes called *flakes, chromium checks, hairline cracks,* or *cooling cracks,* evidence themselves as actual discontinuities in the metal. These are very detrimental conditions and render the steel unfit for use. There is varying opinion as to their exact cause, but in a large class of steels slow cooling will overcome this defect.

Dendrites.—Dendrites are essentially the result of crystallization characteristics of the ingot and are not wholly eliminated in subsequent working. The structure can be partially controlled by the casting temperature and rate of solidification of the ingot. Ordinary heat-treatment will not eliminate the structure. Service failure is apparently little affected, provided that the dendritic segregation is not in the form of nonmetallic inclusions that are incapable of being dispersed and that have been subjected to sufficient mechanical work.

Flow Lines from Forging or Rolling.—A longitudinal section that has been forged or rolled will show streaks or striations, which are known as *flow lines.* They indicate the direction in which the metal has been worked. Flow lines themselves do not signify defective material, since all steel will show flow lines in the direction of elongation. This characteristic may prove valuable in determining whether a piece has been forged, rolled, or cut in shape; it also shows the resultant flow of metal in the section, indicating the position of resultant stresses.

Some of the conditions, developed by macroetching, are shown in Figs. 2-II, 3-II, and 4-II.

Surface Defects.—The common types of surface defects generally developed that may not be visible are seams, laps, and cooling cracks. Seams can be distingiushed by the fact that they are not deep, are elongated in the direction of working, and appear as a thin line. Cooling cracks are most likely to occur in self-hardening steels and can be distinguished by their depth. A lap, as the term implies, is a folding over of the steel. Laps resulting from improper rolling can usually be distinguished from seams by the regularity of their occurrence.

Grinding cracks are generally distinguished by their forming a pattern of symmetry. The fact that the material evidences grinding cracks does not imply that the grinding was improper, since almost any steel if it is overheated in hardening will become so brittle that the most careful grinding will cause these cracks to form.

THE METALLURGICAL MICROSCOPE

For general examination, interpretation, and permanent record of the polished and etched specimens of metals, the metallographist employs the metallurgical microscope. No attempt will be made in the following discussion to consider the problems relating to the optician or the designer, but if we are to use intelligently the tools at our disposal, we must know some of their construction principles and be able to put them to a critical test. Intelligent use, therefore, demands a knowledge of the mechanical and optical units, along with the properly acquired technique.

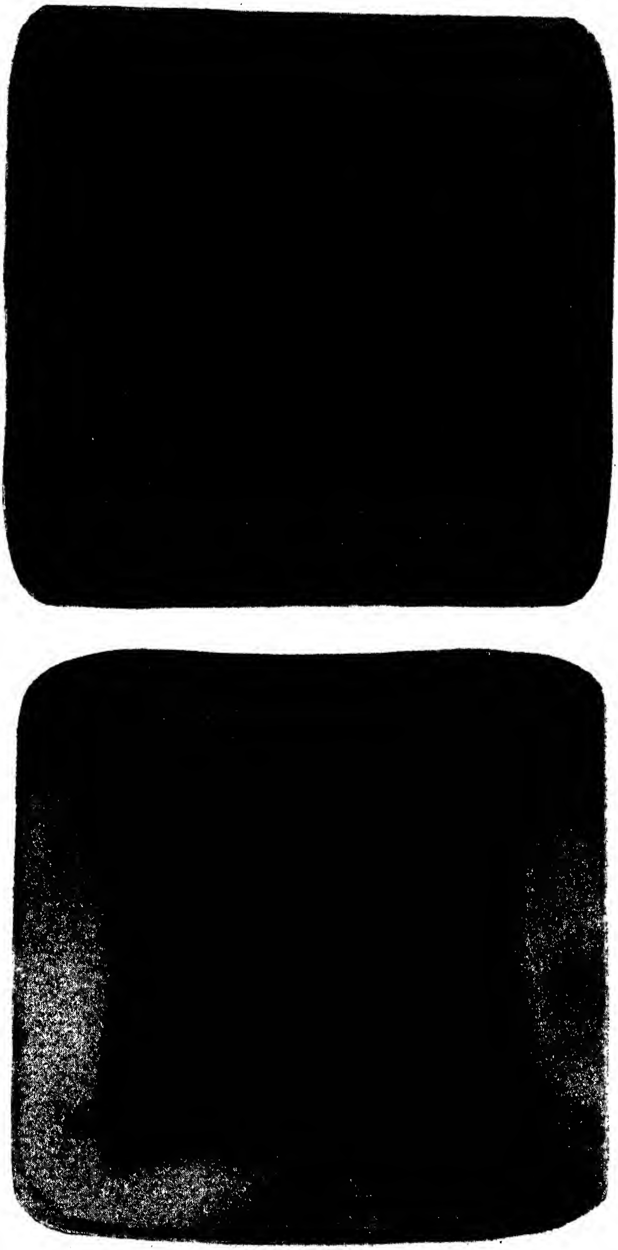

FIG. 2-11.—Photographs of hot acid-etched 4-in. billets. (*Courtesy of Bethlehem Steel Company.*)

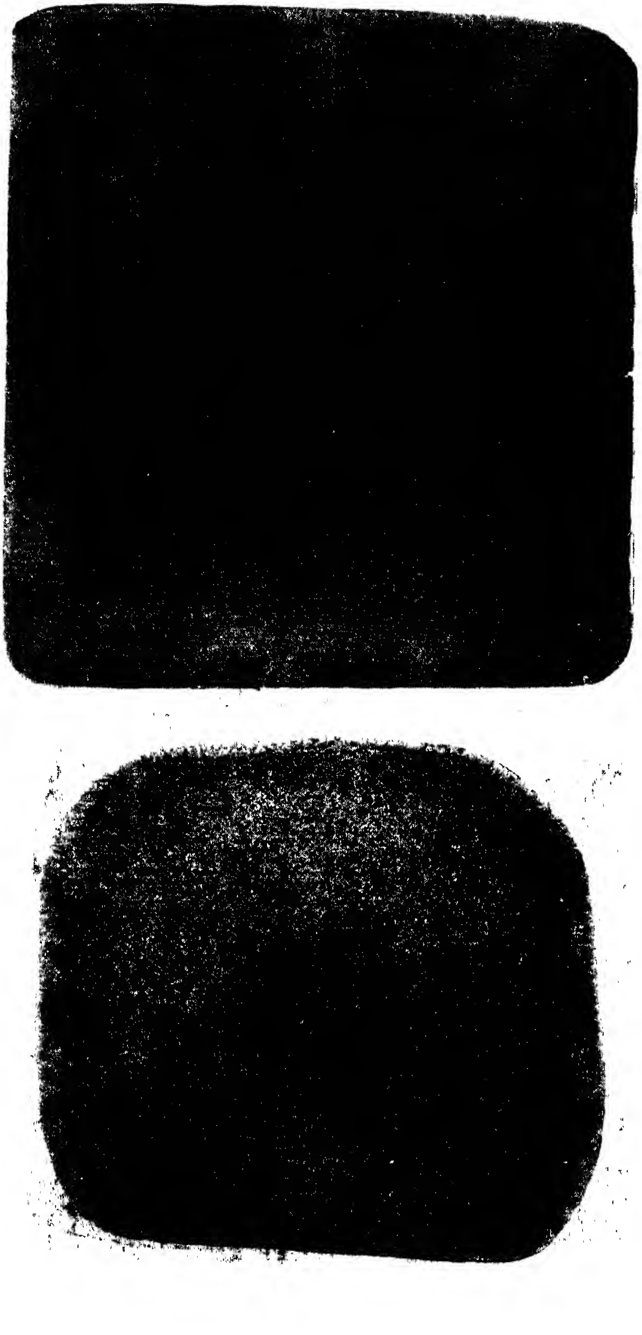

FIG. 3-II.—Photographs of hot acid-etched 4-in. billets. (*Courtesy of Bethlehem Steel Company.*)

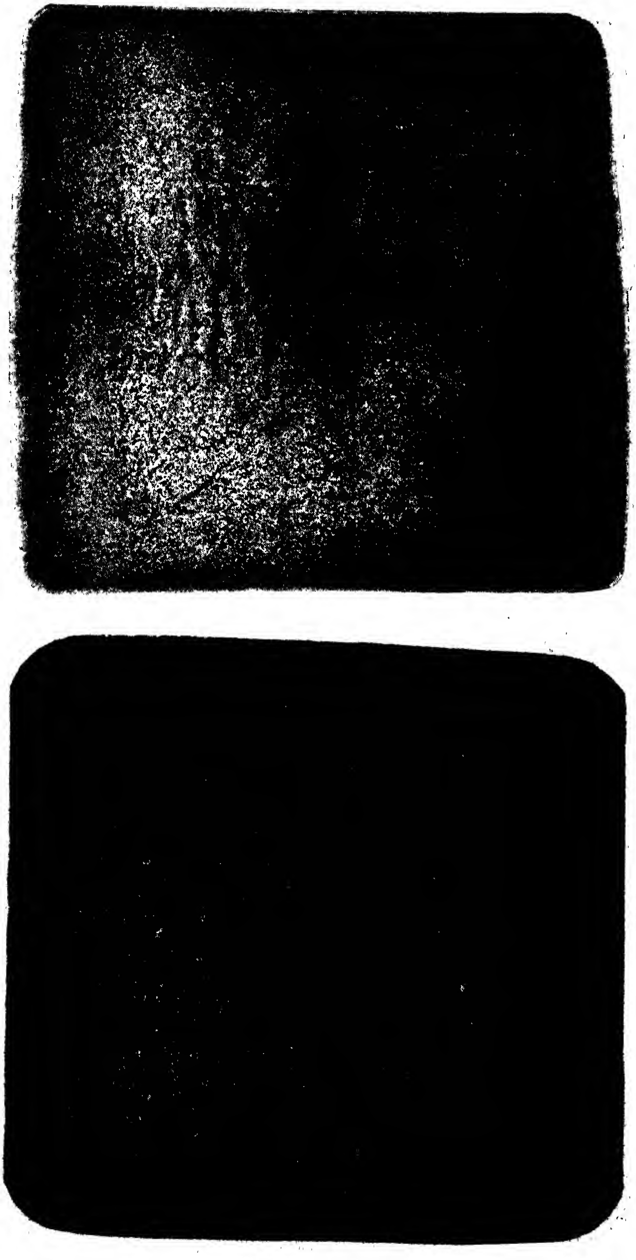

FIG. 4-11.—Photographs of hot acid-etched 4-in. billets. (*Courtesy of Bethlehem Steel Company.*)

With this understanding the metallographist is ready to produce sharp, detailed images at varying magnifications and obtain the full potential resolving ability of the optical system.

For specimen examination, there are many types of metallurgical microscopes that are adaptable to the most exacting research or rapid, routine work. They may be broadly divided into two classes: direct, in which the object is viewed directly through the eyepiece and the objective; and inverted, in which a prism or mirror is placed between the eyepiece and the objective so that the line of light is projected upward, the specimen being placed, with the polished side down, upon a stage directly above the objective.

The Laboratory Microscope.—This type (Fig. 5-II) is adaptable to the examination of a great diversity of material. It is designed to produce satisfactory examination in routine and recording work.

General specifications of this type may be divided into two groups, namely, the optical units and the mechanical stand. The chief optical units are the oculars and objectives, which form the image of the objective, and the illuminating apparatus for the purpose of directing the rays of a source of light toward the object. The mechanical stand consists of a base, which serves as a support; the inclination joint, which serves the purpose of permitting the observer a convenient position; and the handle arm, which acts as a grip for carrying the microscope, supports the body tube, and accommodates within its interior the fine and coarse adjustments.

The body tube of the microscope is of fixed optical length, "tube length," or the distance between the objective and the eyepiece. It is important that each objective used should be employed at the proper tube length for which it has been calibrated. Objects of considerable size may be examined with this setup by means of the adjustable stand and the focusing stage, controlled by coarse and fine adjustments.

A plano-convex mirror in one simultaneous mounting is generally supplied for furnishing understage illumination. For most metallurgical uses the body tube is provided with a vertical illuminator, which serves to direct a horizontal beam of light down through the objective and on to that portion of the specimen under observation.

The body tube, with coarse and fine adjustments, vertical illuminator, and light source, form a very flexible unit, which is adaptable to the examination of specimens both in the field and in the laboratory.

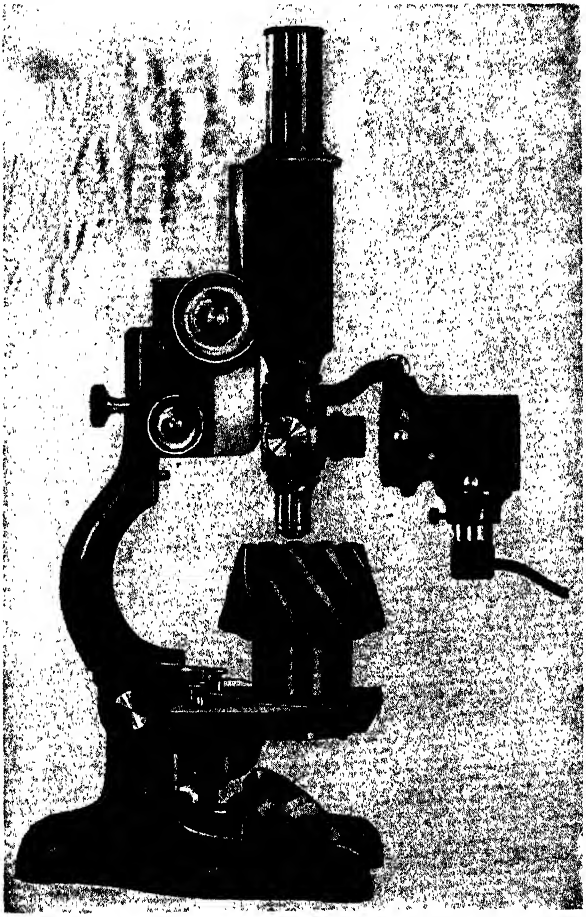

FIG. 5-II.—The utility laboratory microscope. (*Courtesy of Bausch & Lomb Optical Company.*)

Although microscopes are generally equipped with accessories for the examination of opaque or transparent objects in polarized light, we will not attempt an account of this type of examination.

Small cameras may be adapted to the laboratory microscope, the eyepiece merely being pushed into the narrow end of a vertical camera bellows; the microscope projects the image upon a focusing screen, which can be replaced by a plateholder that houses the plates for photographically recording the image. This affords a great flexibility in operations for routine recording and optical work.

Inverted Metallurgical Microscope.—This type of microscope (Fig. 6-II), as originally designed by Le Chatelier, contains the identical features of the direct type, the essential difference being in the mechanical arrangement and direction of vision.

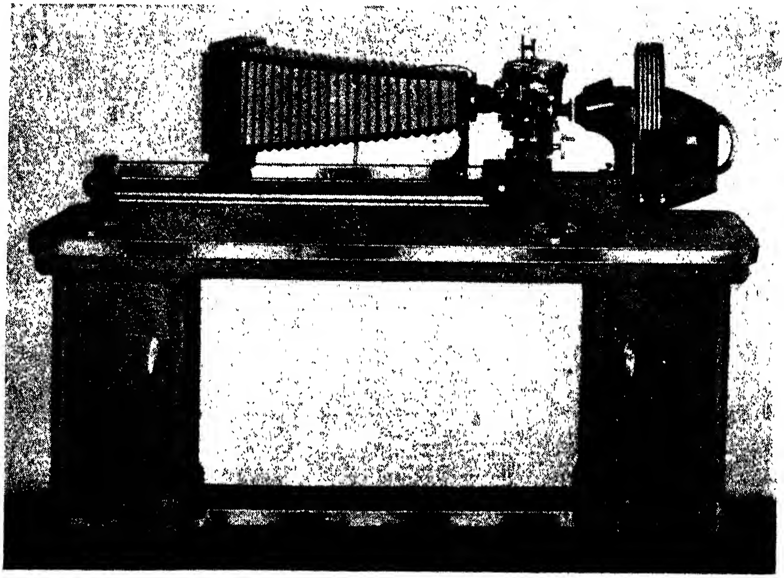

Fig. 6-II.—Research-type metallograph. (*Courtesy of Bausch & Lomb Optical Company.*)

The basic equipment of this type of microscope is as follows (reading from the right of Fig. 6-II): (1) the illuminating apparatus, which consists of an arc lamp with clockwork feed all in a housing, a condenser in front of the arc lamp for focusing the light source, and a porcelain cell for the absorption of objectionable heat radiation; (2) the microscope stand, consisting of a sliding stage, which permits the movement of the specimen in any desired direction, the objective, an illuminator, which may be either plane glass or a prism, and an observation tube at the front of the stand; (3) the camera, fitted with a bellows and with bellows frames, which serve as supports, mounted on the front and rear (the front frame accommodates a lighttight adapter for connection with the microscope stand. A photographic shutter is mounted at the front of the adapter, for convenience in making exposures. The rear frame carries the ground-glass focusing screen and the plateholder); (4) a bench, which is supported on four antivibration shock absorbers, and on which are supported and aligned the stand, the camera and the illuminating apparatus.

Optical System.—The optical system of the microscope consists of two parts: the objective and the eyepiece. The objective, so called because it is always placed nearest the object, is a lens of very short focal length[1] and produces a real, inverted image of the object at some definite plane within the tube. This is a magnified image; the degree of magnification is dependent upon the relative distance from the object to the lens and from the lens to the image; the focal length of the objective determines what these two distances must be and, therefore, the magnification. The image as formed by the objective is examined by means of the eyepiece, which is simply a small magnifier and, in the case of photography, acts as a projection lens.

Defects in Lens Systems.—The question now arises why so many sizes and kinds of lenses are necessary. It is because of the nature and physical properties of light and the natural characteristics of lenses. If we were to employ a simple set of lenses to form the image, we would soon detect many defects that would make the system unsatisfactory. If we take a suitable combination of lenses, we are able to correct these defects for all practical purposes. It is best then that we understand some of these defects so that we may be able to use the proper lens combinations.

White light, as we all know, is a compound of the rays of all colors, each of which has different properties. If this same white light be allowed to pass through a simple convex lens *L* (Fig. 7-II-*A*), the violet rays, which are refracted the most, will be brought to focus at the shortest distance; while the red rays, being

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[1] The principal focus of a lens is the point that is the focus when incident light is a beam parallel to the axis. Focal length is the distance of the principal focus to the lens.

refracted the least, will have the longest focal length; the rays of intermediate colors form a linear spectrum within the extremes. Also, since the focal lengths are different, a screen placed at the point where the violet rays come to a focus would show a circular patch of light, blue in the center and reddish toward the edges, while a screen at the point of focus for the red rays would show a reverse effect. This defect of a lens, which is known as *chromatic aberration* (Fig. 7-II), is sometimes very troublesome and must be corrected in both the objective and the eyepiece of the microscope in order that a sharp image may be obtained.

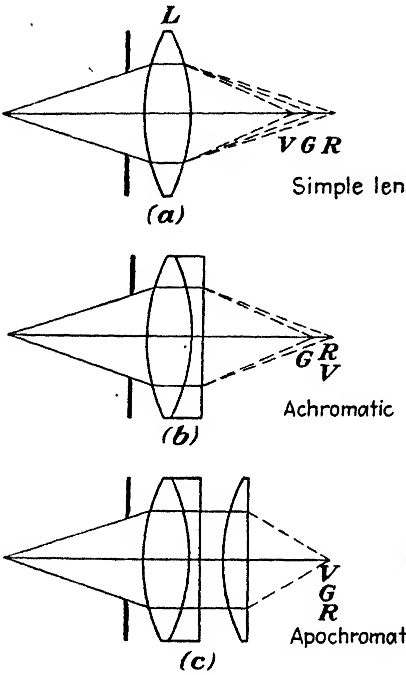

FIG. 7-II.—Chromatic aberration. (Lenses are diaphragmed or stopped down to minimize spherical aberration.)

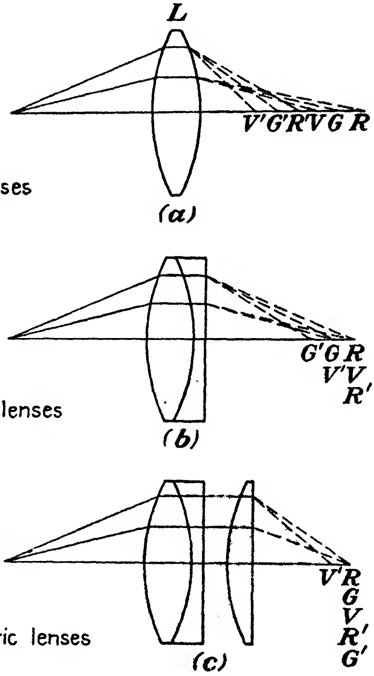

FIG. 8-II.—Chromatic and spherical aberration. (The marginal rays producing spherical aberration are denoted by prime.)

Upon examination of the simple convex lens, we note that the rays of all colors or wave lengths, when passing through the marginal zone of the lens, are focused at a different plane from those passing through the center. This defect in image formation is known as *spherical aberration.* (Fig. 8-II-*A*.) Note that this defect would be apparent even if a single color or wave length of light such as that from a sodium-vapor lamp were used.

It is possible to correct both aberrations to some extent by combining two lenses, one a convex and the other a concave lens, made from different kinds of glass. The second lens will have aberrations opposite to those of the first and will, therefore, neutralize to a great extent the defects of the first lens. With this combination the linear spectrum is folded on itself, the red and violet ends are brought into coincidence (Fig. 7-II-*B*), and likewise the yellow and bluish-green rays are combined at a common point. This converges the rays having the greatest sensibility to the eye (yellow green and greenish blue) in approximately coincident focus. At the same time, the spherical aberration is corrected for one color by a slight modification of the surface ("figuring") in polishing the lens components. This system is called *achromatic* (Fig. 8-II-*B*). It affords strict chromatic correction for two colors (*R* and *V*) and spherical correction for one (*R*).

For ordinary purposes, a lens system that has been achromatized for two colors, red and blue, is not objectionable. If necessary, further corrections may be made by the use of fluorite lenses or by the addition of a third lens, thus achromatizing for three colors (Fig. 7-II-*C*). These lenses may contain one component each of crown glass, flint glass, and fluorite. For further perfection, the shape of the components is modified from truly spherical to correct, at least in part, for the spherical aberration (Fig. 8-II-*C*). This system is known as *apochromatic* (free of color). It affords strict chromatic correction for three colors (*V*, *G*, *R*) and spherical correction for two colors (*R* and *G*). For the finest type of investigation the apochromatic objectives and compensating eyepieces are essential for their fine color, spherical corrections, and flatness of field.

Objectives.—An objective must possess the following qualities: magnifying power, numerical aperture, resolving power, depth of focus or penetrating power, illuminating power, flatness of field, and defining power.

Magnification is usually given in the manufacturer's catalogue in the terms of the combination of a certain objective, eyepiece, and tube length at a predetermined image distance. With the many variables in different instruments and different methods of calculation of constants for lenses by the different makers, it is best to calculate the magnification for a particular outfit being used. Experimentally, accurate determinations may be made by projecting the image of a stage micrometer upon the ground glass and measuring its magnified image by means of a small scale. Micrometers graduated on metal may be obtained for this purpose. The rulings most convenient are 0.1 to 0.01 mm. Charts may then be prepared for any combination of eyepiece and objective with the projection distance as one coordinate and magnification as the other.

If we assume that few, if any, objects are perfectly smooth, then light directed on the object will be dispersed in all directions up to 180 deg. The apparent problem then is to be able by means of lenses to gather and bring to a focus as many of these rays as possible.

If we consider a dry (air) and a wet (oil) immersed objective, we find the fundamental difference is one of resolution. The immersed objective has a greater light-gathering power than a dry lens of corresponding focal length. This light-gathering power is expressed by *numerical aperture.*

Professor Abbe was the first to express this condition and did so by the following formula:

$$\text{N.A.} = n \sin u$$

where n is the refractive index of the medium between the object and objective and u is the semiangle of aperture of the cone of light, reflected from some point, that just fills the aperture of the objective.

In Table 1-II it will be noticed that the numerical aperture for apochromatic objectives is larger than for achromatic objectives of equal focal length, hence such objectives have proportionately greater resolving power.

There is a certain limit, beyond which there is no advantage in increasing magnification. This limit is determined by the *resolving power* of the objective and is defined as the ability to "resolve," or make distinctly visible, the finest detail of the object. Magnification is required to make details visible to the eye but if they are not present in the image no amount of magnification can reveal them.

TABLE 1-II*

Focal length, mm.	Numerical aperture	
	Apochromatic	Achromatic
16	0.30	0.25
8	0.65	0.50
4	0.95	0.85
3	0.95	0.85
2	1.30	1.25

* SAUVEUR, ALBERT, "The Metallography and Heat-treatment of Iron and Steel," McGraw-Hill Book Company, Inc., New York, 1935.

Resolution is dependent upon numerical aperture and the wave length (λ) of light. The larger the numerical aperture, the greater the resolution or the finer the structure that may be resolved and seen. The shorter the wave length, the greater the resolution. The finest detail that the objective can show is equal to $\lambda/2$ N.A., in which λ is the wave length and N.A. the numerical aperture.

A problem will serve to illustrate.[1] For a numerical aperture of 1.00, using green light ($\lambda = 0.00053$ mm.), determine the resolution and the total magnification required to just see the resolvable structure.

$$\text{Resolution} = \frac{0.00053}{2.00} = 0.000265 \text{ mm.}$$

An objective of N.A. = 1.00 can show lines in the object that are 0.000265 mm. apart. If the magnification is 100 × the structure becomes 0.0265 mm. apart. No image so fine as that can be seen with the eye, so we must use an eyepiece to magnify the image further. How much must the eyepiece magnify so that the image becomes comfortably visible? The average resolution of the eye is 1.5 minutes of arc, so how much will the image 0.0265 mm. apart have to be magnified to become visible?

At 10 in. reading distance, 1.5 minutes of arc = 0.11 mm.

$$\frac{0.11}{0.0265} = 4 \text{ times}$$

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[1] "Optical Instruments for Examining and Analyzing Metals," Bausch Lomb Optical Company, Rochester, N.Y., 1931.

Therefore, the structure must be magnified four times to just see this structure, or to magnify the image with an eyepiece for visibility requires a magnification of four times, or a total magnification of 400 times. For any value of numerical aperture, it will be found that the total magnification required to just see the structure (0.11 mm. apart) is 400 × N.A. This factor is known as the visibility factor. A factor of 400 is too low for comfort and one greater than 1,000 is too high for sharp detail.

Depth of focus is the ability of a lens to give sharpness to images of points that are a little in front of or behind the object itself. This quality depends upon the numerical aperture and magnification, being inversely proportional to both. This may be expressed by the formula 1/N.A.; therefore, the higher the magnification, the higher the numerical aperture, the less the depth of focus. In practice, it will be found that if such points do not exceed a certain minute size (0.005 in.), they will appear sharp.

Illuminating power is equal to the square of the numerical aperture of the objectives, and when used at the same magnification may be compared in this manner: *e.g.*, a lens of 0.25 N.A. has an illuminating power of 0.062, while one of 0.50 N.A. has 0.25 or four times as great illumination at the same magnification.

Flatness of field may be defined as the ability of the lens to give sharp definition over the whole field of view. This quality manifests itself when the image of a flat surface is projected upon a screen, the center of the image being sharply focused while the outer position will be blurred, and vice versa. Therefore, the image formed by the objective alone lies on a curved surface, the axis of symmetry of which is the optical axis of the objective.

By using a special eyepiece with a negative lens combination, compensation may be made for the curvature of field, when it is used with proper objectives. Also by increasing the depth of focus, *i.e.*, reducing the used aperture, apparent flatness is obtained but at the expense of resolving power.

Defining power is the ability of a lens to produce images that may be sharply focused. It is dependent upon the finest precision with which both spherical and chromatic aberration has been corrected, the perfect centering of the lenses of the objective, and in the general excellence of mechanical skill.

The Eyepiece.—The second part of the optical system to be considered is the eyepiece, so-called because it is placed nearest the eye. This must not be confused with the projection eyepiece, which is designed for use in photographic work. This latter type is adjustable according to the length of the camera bellows, and enables the observer to maintain the objective at the same distance from the object when changing from visual to photographic work, thus avoiding the necessity of refocusing. The function of the eyepiece is to magnify the image as formed by the objective (see resolution, page 31).

The Huygenian eyepiece, so named after its designer, consists of nonachromatic, plano-convex lenses and a diaphragm for limiting the field of view. The upper lens is known as the eye lens and the lower as the field lens. Huygenian eyepieces are made in a variety of powers, that is the number of times it magnifies the image as formed by the objective, as 4 ×, 5 ×, 7 ×, 10 ×, 15 ×. Different makers use different systems to denote powers and this may lead to confusion if proper precaution is not taken in making a comparison.

The *Hyperplane eyepiece* has a color compensation approximately halfway between the Huygenian eyepiece and the compensating eyepiece, which is described below. The Hyperplane eyepiece is especially designed to produce a flatter image and allows for a larger field of view.

If there are chromatic errors present in the image formed by the objective, they may be eliminated by constructing the eyepiece so that it will have such errors in the opposite sign. In the apochromatic objectives, the chromatic errors of magnification have not been corrected, the red image being larger than the blue, although both have been formed in the same plane. Eyepieces known as *compensating eyepieces* have been designed for the correction of this error. This type of eyepiece should always be used with apochromatic objectives.

Recommended combinations of lens systems that are found convenient for obtaining certain magnifications, may be had from Table 2-II.

Illumination.—Methods of illumination of the specimen constitute a most important and difficult problem; for without a proper source of light and a good illuminating system, satisfactory microscopic work is impossible. Metallographic specimens, being opaque, require the projection of light by some means on their surfaces. This may be accomplished in one of two ways: (1) by directing light obliquely on the specimen or (2) by projecting the light perpendicularly on its surface by means of a vertical illuminator.

TABLE 2-II.—LENSES FOR VARIOUS MAGNIFICATIONS*

Linear magnification in diameters	Objective type	Focal length (approx. mm.)	Ocular type†
½– 2	Photographic tessar	153	None
2– 10	Photographic microtessar	72	None
10– 20	Photographic microtessar	48	None
20– 30	Photographic microtessar	32	None
30– 75	Achromatic	32	
75– 150	Achromatic	16	Hyperplanes (B. & L.);
200– 500	Achromatic or semi-apochromatic (fluorite)	8	Periplanes (Leitz); Ampliplanes (B. & L.);
500	Achromatic or semi-apochromatic	5.50	or Homals (Zeiss). 10 × amplifiers
750– 1,000	Achromatic or semi-apochromatic (fluorite)	4	Hyperplanes (B. & L.);
1,000– 5,000	Oil immersion apochromatic (fluorite)	1.9	Periplanes (Leitz); or Homals (Zeiss).

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* From 1933 "Metals Handbook," and Albert Sauveur, "The Metallography and Heat-treatment of Iron and Steel," McGraw-Hill Book Company, Inc., New York, 1935.

† The magnifying power of the ocular should be such that the desirable size of the projected image can be obtained with a moderate bellows length.

For lenses of the photographic type of more than 48-mm. focal length, diffuse illumination, such as daylight or the projection of a beam of light obliquely on the surface of the specimen, should be used. These lenses are known as *microtessars*, rated so as to give magnifications of from ½ to 30 times, depending upon the bellows draw and the lens. They are used in photographing specimens either in their natural size or slightly magnified.

Objectives with 32-mm. or less focal length are all used with oculars and vertical illumination. Either the plane-glass, mirror, or total-reflecting prism type of illuminators, in which the illuminator is placed between the objective and the eyepiece, may be used.

For best results in photomicrography it is necessary to have critical illumination. "Critical illumination is accomplished by first imaging the light source near the opening of the vertical illuminator by means of a condenser. In this opening is a lens having a focal length such that the source image appears, when viewed from the microscope objective, to come from the plane of the diaphragm of the eyepiece. In other words, the source image is reimaged by the lens in the side of the vertical illuminator, in a position conjugate to the specimen."[1]

Arrangement of the apparatus for critical illumination may be had by referring to any of the optical instrument manufacturers' bulletins or the A.S.M. "Metals Handbook."

The source of illumination becomes extremely important in view of critical illumination. Two important points must, therefore, be kept in mind. (1) Since a sharp image of the light source is formed on the specimen, the source itself must be uniform and free from detail. Otherwise it would superimpose itself on the image of the specimen and confuse the structure. (2) The light source must be properly aligned and at the correct distance for proper imaging.

In view of the many different types of instruments now available to the metallographist, it is difficult—in fact, impossible—to attempt to set up rules as to the proper adjustment for all microscopes. The manufacturers always furnish detailed instructions, and the importance of following these details for the correct adjustment of all parts can hardly be emphasized too strongly. However, the essential parts of the optical system have been described and their functions discussed so that it should not be too difficult for an interested person to bring the microscope into proper adjustment after several trials.

For successful photomicrography there are several prerequisites. One is the proper alignment and adjustment of the apparatus. Focusing may be included in the subject of adjustment. For visual focusing we make use of an average ground-glass screen on which a few scratches have been made. A hand magnifier is carefully focused on the scratches and then is used to focus the microscope until the best detail and definition of the

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[1] "Optical Instruments for Examining and Analyzing Metals," Bausch & Lomb Optical Company, Rochester, N. Y., 1931.

image is obtained in the magnifier. Another prerequisite is stability, ensuring freedom from vibrations.

PHOTOGRAPHIC TECHNIQUE

The production of a photomicrograph has for its purpose the reproduction of the structure observed under the microscope with sufficient clearness to convince an observer that it is a faithful reproduction of that structure. This can be accomplished (1) by polishing in such a manner that the true structure will be developed on etching, (2) by etching to produce the contrast that will give optimum detail in the photographic reproduction, (3) by correct adjustment of the microscope, (4) by using an exposure time that produces an image within the latitude of the film or plate, (5) by development of the film or plate to produce a negative with a contrast that can be reproduced by the printing paper, (6) by the use of a printing paper to obtain a positive print from the negative that will reproduce the details in the negative and show them in the proper tone relations.

It is apparent from the previous discussion that the production of a high-quality photomicrograph not only involves a knowledge of specimen preparation and proper use of the microscope but also a knowledge of photography. In the discussion to follow, only the general features of photography essential to the production of a good photomicrograph will be presented. For complete detail consult the references in the footnote.[1]

Photographic Plates and Films.—Photographic plates or films consist essentially of a transparent base material, such as glass or celluloid, that has been covered on one side—in special cases, on both sides—with a thin coating of emulsion. The emulsion is basically a gelatin in which are suspended fine crystals of silver chloride, bromide, or iodide. When these silver salts are exposed to light, they are affected in some manner that enables a reducing agent, known as a *developer*, to reduce the silver halide to metallic

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[1] "Elementary Photographic Chemistry," Eastman Kodak Company, Rochester, N. Y., 1941; VILELLA, J. R., "Metallography Technique for Steel," A.S.M., Cleveland, Ohio, 1938; NEBLETTE, C. B., "Photography," 4th ed., D. Van Nostrand Company, Inc., New York, 1943; KEHL, G. L., "The Principles of Metallographic Laboratory Practice," 2d ed., McGraw-Hill Book Company, Inc., New York, 1943.

silver. The number of silver-halide grains that are developable is dependent upon the intensity of the light that falls on the plate or film. Thus, when a focused image of an object falls on the plate, the bright areas of the object will render developable a larger number of silver-halide grains than will the darkest areas. When developed, the bright areas will be dense because of the large amount of reduced silver, while in those areas where the silver salts have been affected but little or possibly not at all, the plate will be thin and transparent; these will represent the dark areas of the images. Since the plate after development is a complement of the object, it is known as a *negative*.

Photographic plates and films that are best suited for photomicrography are those that possess the following properties:

Color Sensitivity.—All photographic plates vary in sensitivity toward various sections of the visible spectrum. With respect to this regional sensitivity, they may be divided into three groups: (1) ordinary, which are sensitive to the ultraviolet and blue; (2) orthochromatic, which are sensitive from the blue through the green; and (3) panchromatic, which are sensitive to the entire spectrum and in some cases to the infrared.

Plates or films used in photomicrography should have high sensitivity to yellow and green light; *i.e.*, they must be orthochromatic. This will be apparent if we reconsider the subject of resolution, or ability to reproduce detail, which we found varies inversely with the wave length. In view of this, then, the blue or violet end of the light spectrum should develop the finest details; but unfortunately these rays are difficult to focus, have greater scattering power, and have a low order of reflectivity of metal surfaces, which more than offset the advantages. Therefore, blue light as a filter is seldom used. Experimentation with iron and steel specimens has shown that a combination of the yellow-green rays gives the best results. Therefore, in order photographically to record a sharply focused image, we should use a color filter that will absorb the rays most difficult to focus and allow the ones most sensitive to the eye (yellow-green) to reach the screen or the photographic plate that has been sensitized for this color. Thus we find that detail and contrast may be varied according to the requirements by using a monochromatic filter. The most commonly used filters are the Wratten (*B*) green, (*F*) red, and (*K*) yellow. All photographic plates are sensitive to certain colors, so that nothing will be gained by not using a monochromatic color filter with them.

Latitude.—Latitude is the measure of the ability of a plate or a film to produce a satisfactory negative over a range of different exposure times. A plate that possesses a wide latitude will produce a more satisfactory negative than one with a narrow latitude when the exact exposure time is unknown. Such plates also have the ability to reproduce accurately a greater contrast difference in the specimen. The need for sufficient exposure latitude in plates or films to yield properly exposed negatives is apparent when we consider that the arc light, which is the usual light source in photomicrography, varies considerably in intensity even while the exposure is being made and that the various parts of the structure being photographed vary widely in their reflecting power depending upon the polish, composition, and degree of etching.

Resolving Power.—In our previous discussion of resolution, or the separation of fine detail, it was found to be dependent upon the optical system of the microscope. The fineness of detail in the photomicrograph, however, is dependent upon the nature of the plate or film emulsion. In general, it is found that the finer the grain of the emulsion the smaller is the detail that can be recorded in the photograph.

The types of plates especially suited for photomicrography are either orthochromatics or panchromatics as manufactured by any of the well-known photographic concerns. These plates or film may be obtained with such varying characteristics as extremely high contrast, high available contrast with slow speed, medium contrast with medium speed, and extreme contrast.

Exposure Time.—The correct time of exposure is that which places the brightest and darkest portions of the structure being photographed within the limits of the latitude of the emulsion of the plate or film used. In this way, the bright portion is not overexposed and the darkest is not underexposed. If there is any doubt as to the correct exposure, a simple test may be run by exposing successive portions of a photographic plate to progressively longer exposures. This is accomplished by exposing say 1 in. of the plate for 2 sec., drawing the slide out 1 in. more and exposing for 2 sec. longer, and so on until the entire plate has been exposed. This set of exposures, after proper development of the plate, will indicate the correct exposure time under the conditions of the test.

Plate and Film Development. *Developers.*—The purpose of development is to reveal the invisible image that has been formed on the emulsion of the plate or film during the exposure. The developing solutions used bring about complete reduction of the silver salts that have been affected by the light and, since they are selective in their action, they will reduce only those grains in the emulsion which have been subjected to the light.

All developing solutions have as their basis, chemicals that perform three functions, namely, reducing, activating, and preserving.

The reducing agents used to reduce the silver salts to metallic silver are required to be selective in their action and readily controllable with respect to their rate of reaction. The usual agents are Elon or Metol, pyro, and hydroquinone. These agents alone are not active enough as developers unless they are used with an alkali, such as sodium carbonate. The speed of the development is controlled by the degree of alkalinity and will increase along with the contrast with increase in the alkalinity of the solution to a point where the action will no longer be selective but will reduce all the grains in the emulsion and produce fog over the plate. The action of the developer can be controlled and fogging prevented by the use of a soluble bromide, usually potassium bromide.

A preservative also is required, since developing solutions are quite susceptible to rapid oxidation by the air, by dissolved oxygen in the water, or in the process of development. The usual preservative is sodium sulfite.

The ultimate choice of the developing solution is dependent upon the type of plate and the contrast desired in the negative. It has been found that the contrast in the negative depends upon the type of developer used, the time of development, and the temperature of the developing solution. The contrast will increase with developing time in a given solution, provided that the temperature and the agitation are constant. There is, however, a limit to the contrast obtainable upon prolonged development. In order to obtain negatives of uniform quality the developer formula recommended by the manufacturer of the plate or film should be used.

Fixing.—After the plates have been developed, they should be thoroughly rinsed in water and then placed in a fixing bath. The purpose of this solution is to make the silver halide remaining in the film water soluble so that it can be removed and leave the silver image in the gelatin. Hypo or sodium thiosulfate is the agent almost exclusively used. A hypo solution alone, however, would become discolored by the oxidation of the developer carried over by the negatives and would cause staining of the gelatin. To prevent this and to harden the gelatin, an acid hardening and fixing bath is generally used. Such a bath usually contains a weak acid, such as acetic, to neutralize the developer carried over on the plates; a preservative, such as sodium sulfite; and, a hardener, such as potassium alum, to harden the gelatin.

After fixing for at least twice the time required to clean it, the negative should be washed in running water for 30 to 60 min. and then placed in a drying rack to drain and dry.

Photographic Printing.—Print papers are sheets of paper stock with an emulsion very similar to that on a plate or film, although not so light-sensitive. Such paper, when placed in contact with the negative and exposed to light, will have imprinted upon it a positive image of the negative. This image will then be revealed by development of the paper in a suitable solution.

The reproduction of all the detail in the negative in proper tone qualities depends upon (1) the proper choice of the grade of paper to fit the contrast of the negative, and (2) using the proper exposure time, which will print in detail the densest part of the negative. This can be accomplished by making use of the grades of photographic papers that are available. Papers may be obtained as low contrast, or "soft," for use with high-contrast negatives, and as "hard," or contrasty, for use with flat or soft negatives. Papers may also be obtained in a variety of surface finishes, but it has been found that for photomicrography the glossy is the most satisfactory because it reveals much finer details of the structure.

Development.—After the print has been exposed, it is developed and fixed in solutions that, in general, contain the same ingredients as are used in the development of the negative. Again it should be stressed that for best results the developing and fixing formulas prescribed by the manufacturer should be used.

After fixing, the prints are thoroughly washed and then generally dried on a ferrotype plate, which is a flat plate that has been either black japanned or chromium plated, to impart a high luster to the surface. When dry, the prints will have a brilliant, glossy surface that will show the finest detail.

Suggested Questions for Study and Class Discussion

In Fig. 9-II, *A* is a photomacrograph of a pierced round, which illustrates many of the conditions discussed in the previous text material. *B* and *C* are photomicrographs through a significant section of the round.

A *B* *C*

Fig. 9-II.

1. What sampling and polishing procedures may have been followed in preparing the samples for micro and macro examination?

2. Discuss the various defects shown in *A*.

3. Discuss the material entrapped in the seams, as shown in micros *B* and *C*.

4. What conditions may have been responsible for this seam?

5. Discuss the procedure and the equipment that were probably used in producing the above photographs, placing the greatest emphasis upon the desirable qualities of a good negative and print and telling how these qualities are attained.

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CHAPTER III

THE X RAY, THE GAMMA RAY, AND MAGNETIC TESTING

THE X RAY

In the year 1895, Professor Roentgen of the University of Würzburg, Bavaria, discovered X rays, so named because of their mysterious and unexplained nature. Since that discovery, the greatest strides have been made in the medical application of X rays, as used in the examination and treatment of the human body. It is interesting to note that in Roentgen's paper the first picture made was that of a compass card and needle enclosed in a metal box. The application of the X ray to the examination of metals has been much slower in developing than has the use of these rays for human relief, although metal examination is of great importance. It seems almost as necessary to determine the "ills" of the internal structure of a metal casting, or the essential parts of an automobile or an airplane, as it does to ascertain the ailments of the human body.

It is beyond the scope of this text to discuss deeply the theory of this subject, and so for more detail careful consideration should be given to the references in the footnote.[1]

The X-ray tube most commonly in use in this country, the Coolidge tube, was devised by Dr. W. D. Coolidge of the General Electric Company. This tube is a thin-walled, highly evacuated tube containing a tungsten filament and a block of metal called a *target*, both attached to suitable electrical connections.

When the tungsten filament is heated to incandescence by a low-voltage current, it will emit negatively charged particles known as *electrons*. Now if a high-voltage direct e.m.f. is applied to the tube so that the filament, or *cathode*, is negative, these

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[1] A.S.M. "Metals Handbook," Cleveland, Ohio, 1939; Clark, G. L., "Applied X-rays, 3d ed., McGraw-Hill Book Company, Inc., New York, 1940; "X-rays in Industry," Eastman Kodak Company, Rochester, N.Y., 1930; Industrial Application of the X-ray, *Bulletin* 284, General Electric X-ray Corporation, Schenectady, N.Y.

electrons will pass across the space in the vacuum tube and strike the positive target, or *anode,* at a high velocity. When the electron particle strikes the metallic target, or anode, an *X-ray pulse* is radiated in all directions. With a constant stream of electrons we obtain a continuous stream of pulsations, or a continuous beam of X rays.

The *quantity* of X rays produced with such a tube is dependent entirely upon the temperature of the tungsten filament, which is in turn dependent upon the amount of current applied to the cathode. Similarly the *quality* is dependent upon the velocity with which the electrons are propelled toward the target and upon the voltage applied to the terminals of the tube.

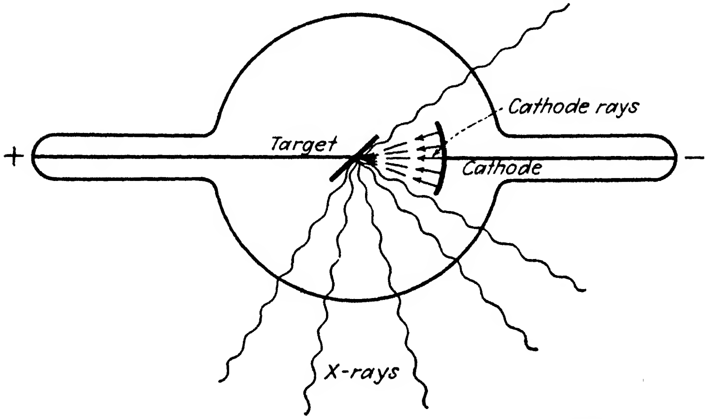

Fig. 1-III.—Graphical representation of an X-ray tube.

In order to utilize the greatest number of electrons emitted, the cathode surface is shaped in the form of a concave mirror, at the center of the curvature of which the target is placed and at which point the electrons are focused. From this focal point the X rays are then emitted, as shown in Fig. 1-III.

What are the *characteristics* of these emitted rays? X rays, like light waves, travel in straight lines, have the same speed as light, and obey the same laws as light, the significant difference being in their wave length (λ). The wave length of the center of the light spectrum is approximately 5.5×10^{-5} cm., while the approximate center of the X-ray spectrum is 5.5×10^{-9} cm. X rays are thus about 1/10,000 of the wave length of light. It is this difference in wave length that enables the X ray to penetrate materials that absorb or reflect the longer wave lengths of light. Further, we find X rays are reflected, scattered, transmitted, absorbed, and reradiated.

Diffraction Methods.—In 1912, Professor von Laue made the discovery that X rays are diffracted by crystals in much the same manner that light is diffracted by a prism or by the finely ruled lines of an optical grating. This discovery and subsequent developments have made it possible for science and industry to determine and control metal structures under given conditions by X ray methods. There are several different methods used for studying the crystalline structure of metals. The three most important are the Laue, the pinhole, and the powder methods.

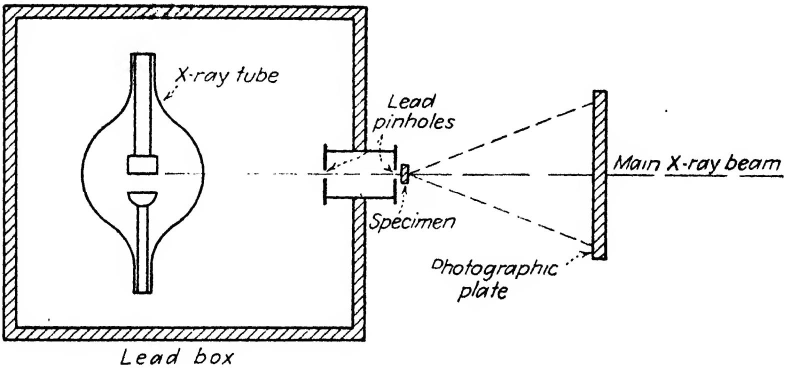

FIG. 2-III.—Graphical representation of the Laue and monochromatic X-ray methods. (*Courtesy of G. L. Clark.*)

The Laue method consists in passing an X-ray beam of many wave lengths through a pinhole, which confines the beam to a narrow pencil of rays, onto a small crystal of the material. The photographic film is mounted at a fixed distance beyond the specimen and perpendicular to the original X-ray beam, as illustrated in Fig. 2-III. The resultant pattern on the photographic plate is a series of spots, each of which represents reflections of a given wave length from a certain crystal plane in the material. Figure 4-III-*A* illustrates the type of pattern obtained from the procedure. This method finds its principal use in the determination of crystal structures.

The powder method consists in passing a monochromatic beam of X rays through a pinhole onto a powdered crystal specimen space between the atoms that affect the diffraction. As a result, there will be quite an intense spot in the center of the photographic film. With proper setup, a typical pattern, such as the one shown for iron (Fig. 4-III-*A*), can be obtained.

What type of pattern will be obtained from using either a solid specimen or a powder specimen? It is known, of course, that all metals are crystalline. Being crystalline, they consist of a great many of these small unit crystal structures. In any one grain of this piece of metal all these tiny units are packed together in a uniform fashion, all the atoms and planes in each unit being parallel to the atoms and planes in every other unit. In a solid specimen or in the case of a powdered specimen there will be a great number of orientations, but in this confusion some of these crystalline units will line up with others and will satisfy the conditions set up by Bragg's law. In this case, even though they satisfy the law, it does not mean that they have to be in the exact position postulated in the discussion of a single crystal. If the reader will consider the unit crystal as being penetrated by a long pin, which would represent the incident X-ray beam, and will rotate the crystal around this pin, it is easily seen that the condition for reflection is satisfied for any position or orientation around the pin, or axis, of the X-ray beam. Thus the diffracted beam will now be transmitted as a

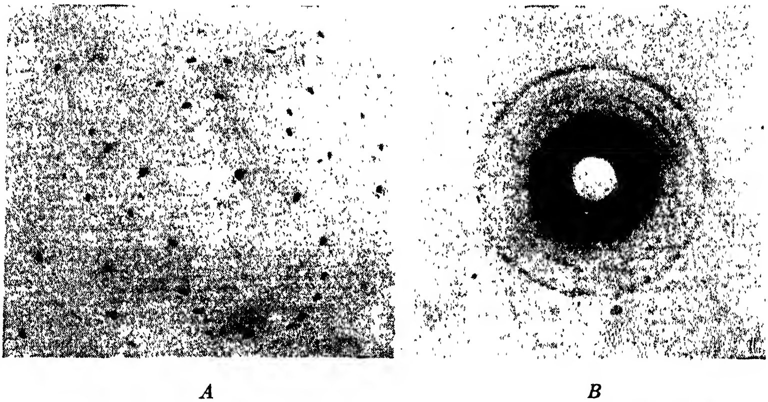

FIG. 4-III.—*A*, Laue pattern of a single crystal of ferrite (containing about 4 per cent silicon), molybdenum radiation. *B*, Debye-Scherrer-Hull transmission photogram through a mild-steel sheet with small grain size. (*Courtesy of Metals Research Laboratory, Carnegie Institute of Technology.*)

circular cone and, if the record were made for each possible reflection upon a photographic plate mounted perpendicular to the incident beam, there would appear a series of rings surrounding the central spot. Figure 4-III-*B* shows such a pinhole diagram, while Fig. 5-III is a powder diffraction pattern taken on a narrow film that has been placed in the circumference of a circle at the center of which is the specimen. A horizontal section cut through Fig. 4-III-*B* will produce the line spectra of Fig. 5-III.

Interpretation.—Now that certain diffraction effects have been observed, particularly for iron, the interpretation, which is the next step, demands chiefly knowledge and experience. There will be no discussion of the problem involving a complete structural analysis, which means a determination of such things as crystallographic system—*i.e.*, whether it is cubic, tetragonal, hexagonal, etc.; the volume occupied by the unit crystal; the number of atoms or molecules in each cell; the symmetry of the molecule; the positions of the diffracting centers; etc. However, the interpretation of the condition of the substance of a known structure in terms of grain size and orientation, internal strains or distortion, extent of deformation, effect of heat-treatment, etc., which may be obtained from the diffraction patterns and which are most important from an industrial point of view, will be discussed briefly.

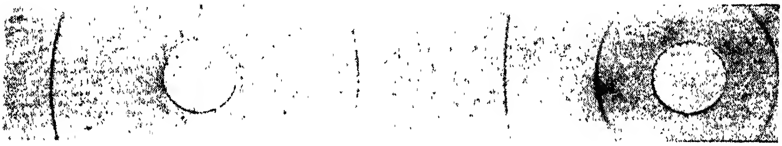

FIG. 5-III.—Powder photogram of iron taken with iron radiation (small Debye camera; holes punched in film for incident and transmitted X-ray beam). (*Courtesy of Metals Research Laboratory, Carnegie Institute of Technology.*)

Grain Size.—Figures 4-III-*A*, 7-III-*A*, and 4-III-*B* are Laue diffraction patterns that show the effect of variable grain size in a metal. The first pattern shows the effect of passing the X-ray beam through a single crystal, while the succeeding patterns show the effect when increasing numbers of grains are present, up to one million or more, in a random state of orientation. The first pattern shows an unsymmetrical array of Laue spots, while with an increasing number of grains the spots become more numerous and form a series of concentric, continuous rings (Debye-Scherrer rings). All the patterns shown are characteristic of alpha iron and, in addition, each characterizes a particular condition of grain size. There may be every possible gradation of grain size between the extremes and extending, besides, to grain sizes in the colloidal range smaller than that shown in the last pattern of Fig. 4-III-*B*.

Orientation.—The diagrammatic structure of the sheet metal with the grains randomly oriented is shown in Fig. 6-III-*A*. In severely deformed objects, such as wire or sheets, the grains are found not to be randomly oriented, but to line up their crystallographic planes in positions parallel and transverse to the direction of rolling. Figure 6-III-*B* shows the diagrammatic structure of the sheet metal with the grains all oriented in a preferred direction with respect to the rolling. In Fig. 7-III-*A*, *B*, *C*, and *D* are shown pinhole photograms of the inside of a mild-steel sheet, all from the same steel, in which, respectively, there was no cold reduction, 60 per cent cold reduction, 80 per cent cold reduction, and 95 per cent cold reduction. Figure 7-III-*A* shows the original grain size of the sheet, which by actual count was about 400 grains per sq. mm. (A.S.T.M. 5-6). From an examination of these photograms it appears that with increasing amounts of cold reduction there are more and more crystal grains contributing reflection effects in certain directions, while in other positions few, if any, grains are available for reflection. This effect is evidenced by the blackening of the film in a series of symmetrical segments. Such patterns show preferred orientation and are generally designated as fiber diagrams. The preferred orientation, in these figures, is gradually developed by the increased amount of cold work and approaches quite closely to a condition of complete orderliness of orientation of all the crystals.

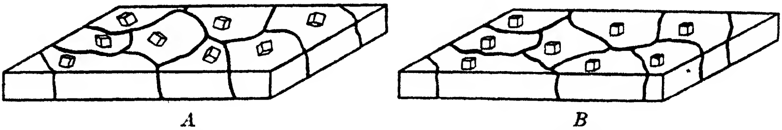

FIG. 6-III.—*A*, Structure of sheet metal with randomly oriented grains. *B*, Structure of rolled sheet (metal) steel with preferred orientation of grains. (*After G. E. Doan.*)

The texture of the sheet gives rise, in these cases, to four-point diagrams, evidenced by the blackening in four segments. This condition, according to the recent works of Gensamer and Mehl

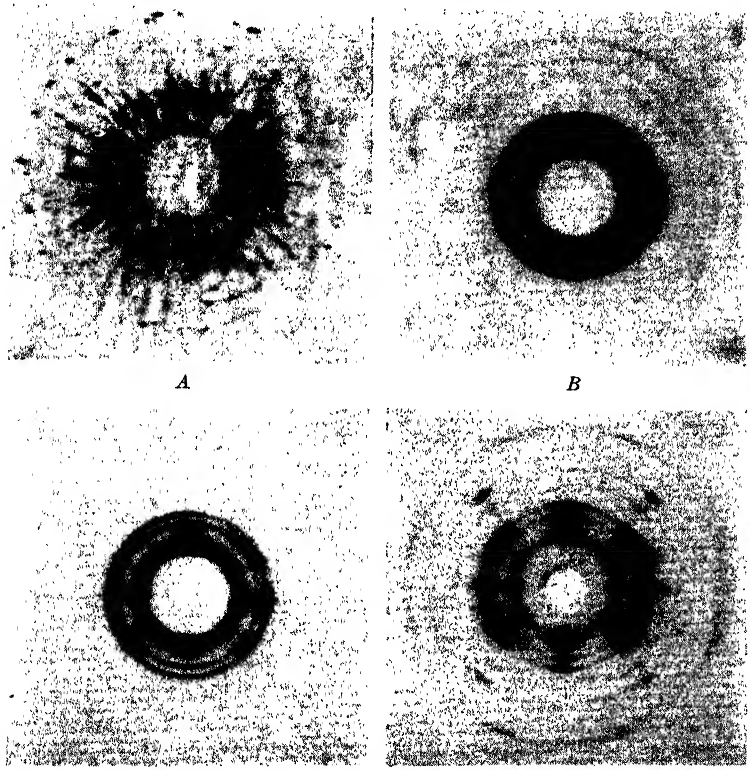

FIG. 7-III.—*A*, Laue pattern of a sample of mild steel normalized four times. The resultant grain size is fairly large (A.S.T.M. No. 5-6). Beam 90 deg. to surface, sample reduced 0 per cent. *B*, Fiber pattern of the steel shown in Fig. 5-III. Beam 90 deg. to surface, sample reduced 60 per cent, inside of sheet. *C*, Fiber pattern of the steel shown in Fig. 5-III. Beam 90 deg. to surface, sample reduced 80 per cent, inside of sheet. *D*, Fiber pattern of the steel shown in Fig. 5-III. Beam 90 deg. to surface, sample reduced 95 per cent, inside of sheet. (*Courtesy of Metals Research Laboratory, Carnegie Institute of Technology.*)

is the typical pattern to be expected from the inside of a cold-rolled mild-steel sheet when the X-ray beam is perpendicular to the direction of rolling. If the beam had been parallel to the direction of rolling, an entirely different fiber pattern would have been produced. In wire the same pattern would be obtained with any orientation of the specimen with respect to the beam, as long as the beam passed perpendicular to the wire axis.

Internal Strain.—Strain in metal structures may be induced in many different ways, a number of which will be discussed later in the text. The detection and even the quantitative estimation of internal strain can be made by X-ray diffraction methods. Figure 8-III shows what happens to the Laue diffraction pattern of the single crystal of iron, previously shown, when the crystal has been severely distorted by bending. The spots are now elongated to radial streaks or "asterism streaks." Any such pattern, even for a polycrystalline material that shows these streaks, is evidence of internal strain. The crystal planes are now distorted so that the reflections take place as from cylindrical mirrors instead of plane mirrors.

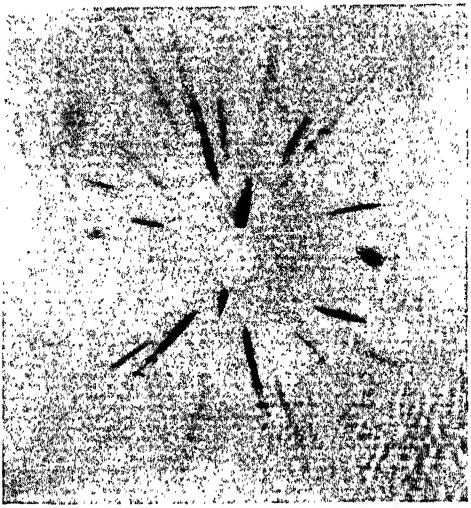

FIG. 8-III.—Laue pattern of a single crystal of ferrite, as in Fig. 4-III-*A*, severely distorted by bending. The resultant strain is shown by the asterism streaks. (*Courtesy of Metals Research Laboratory, Carnegie Institute of Technology.*)

Recrystallization of Cold-worked Metals.—In the previous section the result of cold deformation in bringing about strongly directional properties was shown. Commercially, this condition, with several exceptions, is undesirable, so the recourse is to heat-treat and bring about a recrystallization of the grains while retaining the form of the sheet or wire. Heat-treatment, as would be suspected, produces a random, nondirectional orientation. Figure 9-III shows the pattern produced following recrystallization of the steel, which had been cold-reduced 95 per cent, at 577°C. (1068°F.). The sharply localized spots denoting preferred orientation in the rolled condition have now become indistinct, with the pattern denoting random orientation predominant. The last traces of fibering are quite persistent and an entirely random arrangement is attained only by heating to very high temperatures.

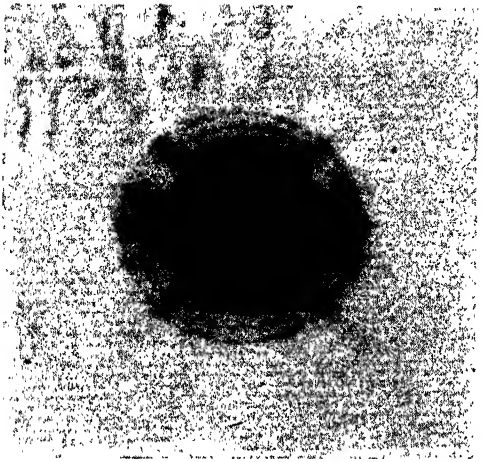

Fig. 9-III.—Laue pattern showing effect of heating to 577°C. (1068°F.) on the pattern of a sheet sample cold reduced 95 per cent Fig. 7-III. (*Courtesy of Metals Research Laboratory, Carnegie Institute of Technology.*)

The X-ray method, in this case, is of practical importance in determining the completeness of recrystallization of the fiber structure and the elimination of internal strain. The elimination of strains is always brought about by heat-treatment.

Radiography.—Now, let us discuss a further use of the X ray, that of the nondestructive examination of the interior of an opaque object. To do this we make use of the property of absorption, or rather of differential absorption, which makes possible the taking of shadow pictures or radiographs.

A thick object will absorb more rays than a thin one and in material of equal thickness the absorption will increase with the atomic weight of the material. Thus, iron will permit the transmission of less energy than will aluminum.

If we take an object that is solid and homogeneous and expose it to the X-ray beam, the radiograph obtained will be of uniform density, while if we take, instead, an object such as a casting that may have an area of nonhomogeneity, such as a crack, slag, or gas cavity, which has less resistance to the X rays, on the radiograph this area will appear denser than the surrounding portion of the film, because more X rays have been transmitted by the object at this point.

To obtain an undistorted, well-defined radiograph it becomes necessary to keep the film as close as possible to the object, to keep the source of radiation as far from the object as is justified with required exposure time, and to align the X-ray beam, the object, and the film so that the shadow image will be as near the original size of the object as possible. Figure 10-III shows a diagrammatic view of this method of testing.

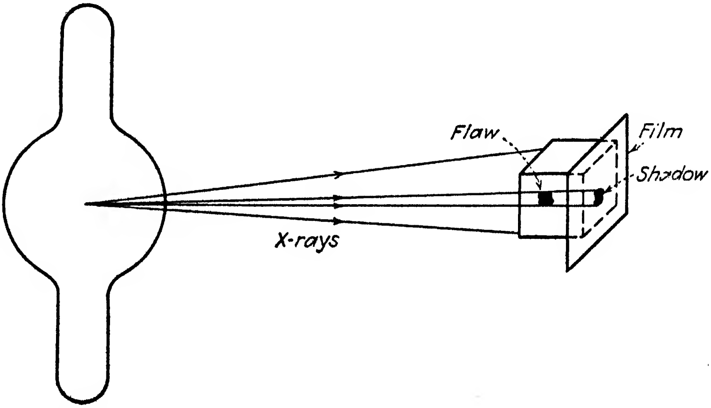

FIG. 10-III[1].—Graphic representation of X-ray radiographic testing.

X rays have a further property—the ability to affect the silver compound or emulsion of an X-ray film. When the X ray strikes the film, which during exposure is protected by a lighttight holder, less than 1 per cent of its energy is absorbed and 99 per cent is transmitted through, without performing any useful work. It is this small percentage that forms the image. For-

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[1] DOAN and MAHLA, "The Principles of Physical Metallurgy," 2d. ed., p. 337, McGraw-Hill Book Company, Inc., New York, 1941.

tunately, however, certain chemicals have the property of absorbing X rays and converting them into ordinary light. Calcium tungstate is one of the chemicals that have this property, which is known as *fluorescence.* Therefore, by making a screen of this ingredient and placing it in contact with the emulsion of the film we are able to make use of a portion of the heretofore wasted rays. The film absorbs a small portion of the rays, while the screen absorbs the greater number and converts them into ordinary light, so that the negative image is formed mainly by

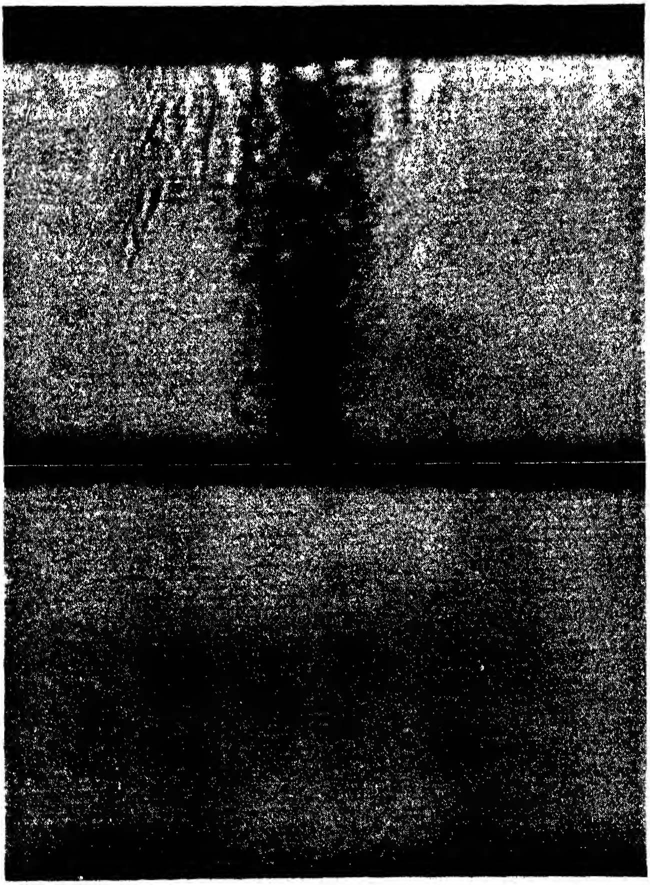

Fig. 11-III.—Radiographs of two welds. One shows a poor weld (top) and the other (bottom) a perfect weld. (*Courtesy of General Electric X-Ray Corporation.*)

the converted energy. Ultimately, the time of exposure is reduced to one-eighth the time necessary without screens.

In making a radiograph it becomes necessary to consider the material, its density and thickness, the tube distance, the affected current and voltage, and exposure time under these conditions. Further consideration must be given to the sensitivity of the equipment and the limiting size of defect in the object that may be detected; the maximum thickness of material that can be penetrated; proper logging and marking of object and film; developing technique; and the proper interpretation of markings on the film.

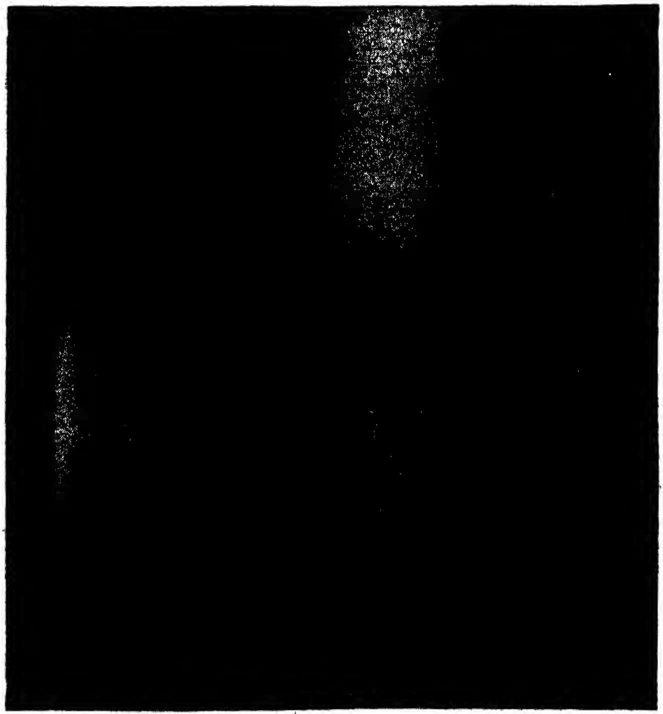

Fig. 12-III.—Radiograph of an aluminum alloy steel casting showing cavity in the vicinity of the gate.

Figure 11-III shows the radiograph of two welds, the markings indicating that one is a poor weld and the other a perfect weld. Figures 12-III and 13-III show, respectively, a radiograph and a

photograph of an aluminum alloy sand casting. The cavity in the vicinity of the gate can easily be seen in both the radiograph and the photograph of the sectioned specimen. Note that a portion of this cavity has been filled with a metal after removal of the gate.

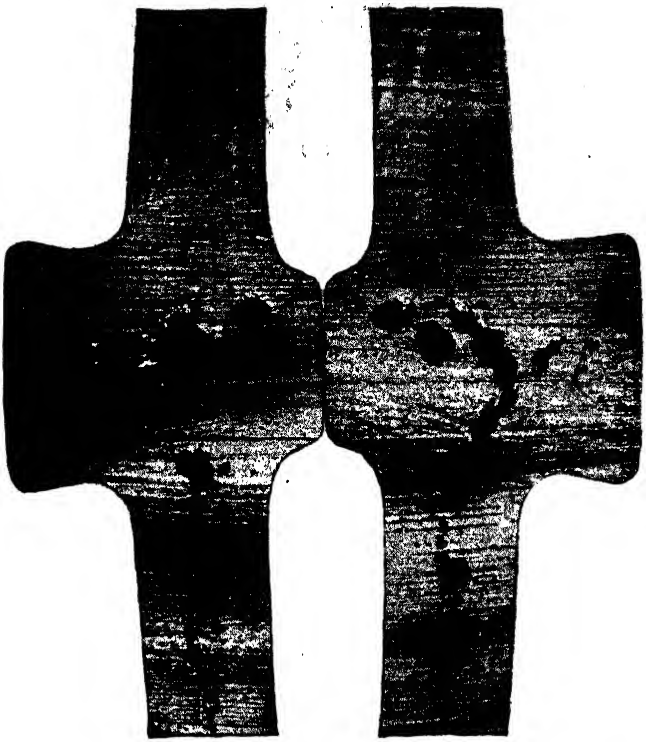

FIG. 13-III.—Photograph of surface after Fig. 12-III was sectioned. (*Courtesy of General Electric X-Ray Corporation.*)

THE GAMMA RAY

The X-ray and the gamma-ray method of nondestructive testing involve practically the same principles, as can be seen by comparing the diagrammatic sketch in Fig. 14-III with that in Fig. 10-III. There is no fundamental difference in the nature of the two rays used except that the gamma ray is much shorter in wave length. The waves coming from a gamma-ray source are not monochromatic; *i.e.*, they do not possess a single wave but rather a band of wave lengths. The effective wave length of a 250,000-volt X-ray tube varies from 0.10 to 0.25 angstrom units (= Å = A.U. = 10^{-8} cm.), whereas the wave length of the gamma ray varies from approximately 0.004 to 0.034Å. The efficiency, or the time in which the rays penetrate a certain material, depends mostly on the wave length of the rays. The shorter the wave length, the harder the ray, the more rapid will be the penetration.

Up to a thickness of 3 or 3½ in., under certain conditions, the X ray may be used more efficiently than the gamma ray for radiographic work; but for thicker materials, because of the long exposure time required, it becomes impractical to use the X ray, whereas with the gamma ray satisfactory pictures have been made of castings 10 in. in thickness. Thus for 6 in. of steel, the exposure time with a 220,000-volt X-ray tube using 5 ma., is 8,000 hr. Using 1000 mg. of radium, at the same distance, the exposure time would be approximately 100 hr. or less.

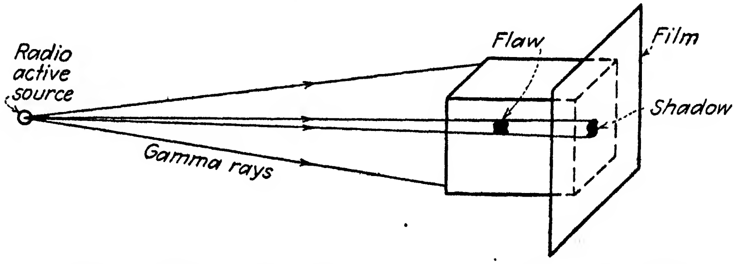

Fig. 14-III.—Graphic representation of gamma-ray radiographic testing.

These short rays are the product of the decomposition of radioactive substances, radium, or radon, the first product of the decomposition of radium. Radium decomposes to chemically and physically different materials, the final product being lead. Transformations are accompanied by radiations, of which there are three types: the alpha rays, which are known to be helium atoms; the beta rays, which are electrons; and the gamma rays, which are "light" waves of extremely short wave lengths.

In order satisfactorily to radiograph material with radium or radon, we make use of the properties of these short "light" rays as we have done in using the short rays emitted by the X ray. Gamma rays (Fig. 14-III) are reflected, scattered, transmitted, absorbed, and reradiated in the same way as X rays, and so the same precautions are necessary. The technique merely consists of aligning the specimen with the radioactive source, placing a holder containing a calcium tungstate screen as an intensifier with photographic film at the rear of the object, retrieving the film holder at the end of the proper exposure time, and developing the film.

Applications of Radiography in the Metal Industry.—1. Development of good fabricating or casting technique and practice for metal products such as sand castings, permanent mold castings, die castings, forging, wire, and assemblies.

2. Partial production inspection.
3. Complete production inspection involving:
   - *a.* High temperature and pressure castings for steam-electric installations, hydroelectric plants, boilers, oil cracking systems, etc.
   - *b.* Welded joints for class I pressure vessels, accepted by the A.S.M.E. boiler code, provided that the welds are completely examined by the X ray.
   - *c.* Highly stressed castings for aircraft and automotive applications such as propellers, connecting rods, and engine parts where production is small and a forging die is too expensive.

For further information see the references in the footnote.[1]

MAGNETIC TESTING

Various forms of magnetic analysis have been in use or in the process of development for the last 25 years. All the various forms of magnetic analysis are based on the phenomena of induced magnetism. Each form of magnetic analysis measures some characteristic or group of characteristics of this induced magnetism. These characteristics are then correlated to the physical and structural characteristics of the steel.

If later on these characteristics reappear for some other sample of steel aside from the standard, then the physical and structural

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[1] Mehl, R. F., Nondestructive Testing by Gamma Rays from Radium, *J. Am. Soc. Naval Eng.*, **43**, 371–395 (1931); Barrett, Gazelius, and Mehl, The Technique of Radiography and Gamma Ray, *Metals and Alloys*, **1**, 872–879 (1930); Doan, G. E., Practical Procedure for Gamma-ray Testing, *Metal Progress*, **25** (No. 5) 22–26 (1934).

characteristics of this steel should be the same as for the standard. It has been experimentally proved by many investigators that the magnetic properties of a material are a peculiar condition dependent on the physical and structural conditions of that material. Also, for this condition only one set or group of magnetic characteristics is possible. It readily follows, then, that if the magnetic characteristics are the same for two specimens of steel these specimens should be in the same physical condition.

In the comparison method of magnetic analysis, the procedure is either to compare the properties of one part of a specimen with another part of the same specimen in order to establish uniformity, or to compare the magnetic properties of a specimen of unknown characteristics with the magnetic properties of a standard specimen or set of specimens whose physical properties have been previously determined. The latter method is preferable for most purposes. This type of test may be used to inspect welds so as to determine their porosity and strength; to determine cracks, seams, nonmetallic inclusions, segregations, internal and external defects; to determine heat-treatment, hardness, variations in microstructure, soft spots, shallow hardness, and grain size; and to separate shapes of the same form but of different composition.

Briefly, this principle involves a comparison of the magnetic permeabilities of a standard sample with the material being tested.

$$\text{Permeability} = \frac{\text{flux density}}{\text{magnetizing force}}$$

For example, within certain limits an increase in the carbon content of steel produces an increase of hardness and tensile strength, decreases the permeability and flux density, but increases the coercive or magnetizing force and hysteresis losses. An increase in the air gap of the magnetic circuit due to minute cracks or gas pockets will readily change the mechanical and magnetic properties of the steel. Thus, it is possible to test for any characteristic in steel that affects its magnetic properties and, thereby, its physical properties.

It must be noted, however, that specimens having like magnetic properties will have like mechanical properties, but the opposite effect is not necessarily true. There are so many variables that affect the magnetic and mechanical property relationship that each application must be worked out independently and the procedure used must be carefully coordinated with the mechanical properties.

Magnetic analysis may also be used for the detection of hidden defects in ferromagnetic materials in a manner other than that just described. The method referred to is the magnetic power method (magnaflux), which detects surface seams immediately below the surface, grinding cracks, quenching cracks, shrinkage cavities in welds, and incipient fatigue cracks in used machinery. In this method magnetic powder, either finely divided iron powder or magnetic iron oxide, wet or dry, is deposited upon the surface of the magnetized piece. The cracks or flaws are revealed by characteristic patterns on the surface. Surface defects are quite easily detected. Subsurface defects are more difficult; their location is dependent upon their size, location, and depth below the surface, and upon the condition of the material and the procedure used. The depth to which subsurface defects can be detected is approximately the same as the cross-sectional height of the defect. The largest problem in the detection of defects by this method is the evaluation of the seriousness of the defect, since different currents and methods give patterns of different intensities.

Magnetic analysis of these types has a variety of uses for which no other type of nondestructive test can be used to give a 100 per cent inspection as efficiently and at such small cost. For more complete details regarding the methods and procedures involved in magnetic testing see the references in the footnote.[1]

Suggested Questions for Study or Class Discussion

1. What defects may be detected by radiographic examination in wrought, cast, or finished steel products?

2. What circumstances would limit the use of each of the three nondestructive test methods?

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[1] Sperry, E. A., Nondestructive Testing, *Trans. A.S.S.T.*, **16**, 771–798 (1929); Kouwenhoven and Seletzky, The Unbalanced A.C. Bridge for Magnetic Analysis, *A.S.T.M. Proc.*, **30**, Part II, 298–312 (1930); Kouwenhoven and Tebo, The Incremental Permeability Method for the Magnetic Analysis of High-speed Steel, *A.S.T.M. Proc.*, **28**, Part II, 356–374 (1928); The Ferrous Magnetic Corporation, Catalogue 2, New York, N.Y.

3. From an industrial viewpoint, what values may be had from a diffraction or crystalline structure investigation?

4. How would an exposure chart (time versus thickness) for a given material be developed for use in radiographic investigation?

5. How might any one or all of the nondestructive test methods be employed in your plant?

6. Since the rays emanating from both the X ray and radium have an accumulative harmful effect on the human body, what care should be taken to protect the operator during the exposure?

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CHAPTER IV

THE PHYSICAL TESTING OF METALS

The most absolute and conclusive test for ascertaining the physical properties of metals is to place them in actual use. For example, the engineer may take an automobile—say, one of this year's models—and operate it continuously day and night for as much as 100,000 miles over all types of roads and under conditions far worse than would ever be encountered under normal conditions. Further, a similar model may be rigged up in the laboratory and tested for horsepower developed by attaching suitable testing equipment to the rear wheels; and to determine how much of a load it has withstood under certain conditions and how much of a carload the springs will carry in passing over a bump or when overloaded. Through the years of development, this knowledge and experience in actual use has been carefully accumulated and recorded in handbooks and literature until now the designer can tell just how strong he must make a certain part, *i.e.*, the load it must carry in pounds per square inch to withstand the most extreme conditions.

This method of testing, although it is the only conclusive method, is time consuming and quite expensive and has as a result brought about the development of a number of rapid methods of testing. Thus, simple tests have been devised to test the approximate suitability of certain metals for certain parts before they are actually made; *i.e.*, the steel is checked to see that it comes within certain specifications. These specifications, having been decided upon by the engineer through his accumulated knowledge of actual service conditions, enable him to make a correlation of these conditions with the results of the simple tests and ensure to a certain degree that the part so made will withstand the demands made upon it.

Further to analyze conditions, the engineer must classify his forces, *i.e.*, the forces that the steel is subjected to and must overcome in actual service. In testing, therefore, we must consider directional force and application of force. Classes of forces with respect to direction are those of (1) tension, forces tending to put the material under tension; (2) compression, forces tending to compress the material; (3) torsion, forces tending to twist the material; (4) shear, forces tending to cut the material across its section; and (5) some combination of these forces. The classification of forces or stresses with respect to application is as follows: (1) static, the result of gradual application of a steady force; and, (2) dynamic, the result of repeated impact or vibratory forces, which may be further subdivided as (*a*) fatigue, the result of repeated application of a force, and (*b*) impact, the result of sudden application of force.

To the tests for forces may be added a number of other tests, such as those for hardness, the resistance to penetration; wear, relative resistance to abrasion or penetration; corrosion, resistance to chemical deterioration when subjected to various agents; and high-temperature properties, involving strength, creep, and resistance to oxidation at specified temperature.

In the following material we shall briefly describe the principal forms of mechanical testing and essential experimental details.

TENSILE TEST

It is difficult to say which of the many physical tests have the greatest commercial importance. The most commonly determined physical test on metals is the tensile test, which evaluates the strength of the metal. Tension, as the term denotes, is the application of a force that causes an extension or stretching. The tensile strength or ultimate tensile strength is the maximum resistance of the metal to the stretching force.

In the process of obtaining the tensile strength of the material other important information on its load-carrying ability is obtained, such as, the elastic limit, the proportional limit, yield point, and yield strength. The yield strength is of particular importance, since it is a measure of the load that a material can withstand without taking a permanent set upon application and release of the load. It is obvious that, should the forces applied to a given structure exceed this figure and take a permanent set, it would after that be of very questionable service value. In view of these facts, the engineer bases his design on the yield strength and a proper allowance for the service conditions by the use of a factor of safety.

In this country the tensile strength or ultimate tensile strength —which denotes the load causing complete failure—and the elastic limit, proportional limit, yield point, or yield strength—which denote the elastic characteristics—are all calculated from the information derived in testing to give the stresses in pounds per square inch.

In addition to the measurements already mentioned, record is made also, after the test specimen has been fractured, of the change in length and in cross section. The deformation produced in the testing to fracture is a measure of the ductility of the material, or its ability to deform. The actual evaluation of this property is a subject of considerable discussion. However, as far as the design engineer is concerned, the specifications are still included with orders for materials for given applications, on the basis that certain specifications have given good service conditions in past experience and, if duplicated, should continue to be satisfactory.

These two measures of ductility, the elongation and the reduction of area (cross-section contraction) are measured after fracturing the test specimen. The calculations are then made with respect to the original dimensions of the test specimen and reported as percentages.

The differentiation of these terms along with the proper calculations of the test results will be discussed in the following sections.

Testing Machines.—The machines used for tensile testing are varied in form and size. All the machines, however, work on one common principle, which involves the application of the force to the test piece by one member, the take up of the extension of the test piece that occurs during the application of the stress by another member, and an apparatus for measuring the stress applied to the test piece. Broadly speaking, there are two main types of machines, embodying two distinct principles—one, operated by direct counterpoise through a system of one or more levers and weights, and the other, operated by hydraulic pressure.

The first type of machine, the loaded-beam type (Fig. 1-IV), is quite rugged, powerful, and easy to operate. It is generally gear-driven and hence it is noisy and causes considerable vibration, both from the machine and from the jolts accompanying the rupture of the specimen. Further, it has a variable set of grips necessary for testing a wide range of specimens, screw posts for application of the load, knife edges for connecting shackles with the lever arm or calibrated scale beam upon which the load is accurately measured.

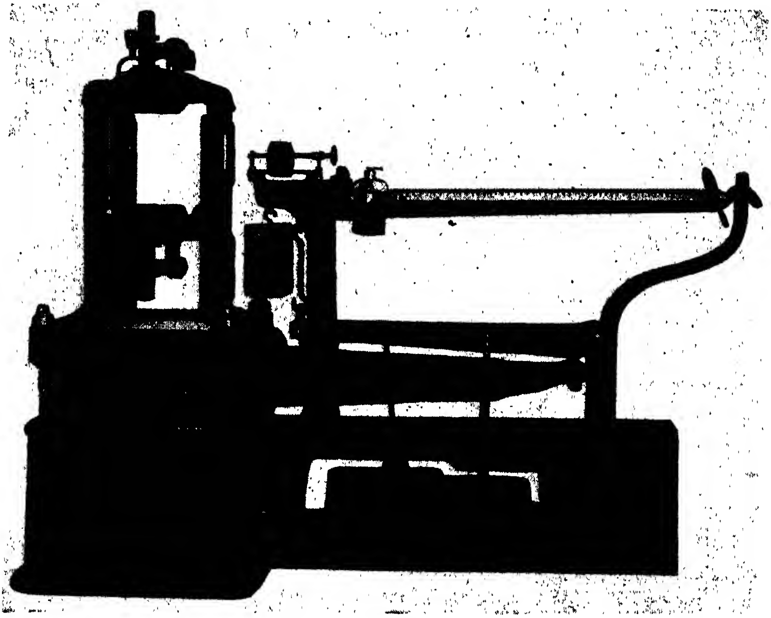

FIG. 1-IV.—Photograph of a loaded beam-type testing machine. (*Courtesy of American Machine and Metal Manufacturing Company, Riehle Division.*)

The second type of machine, the hydraulic type (Fig. 2-IV), is very much lighter than the loaded-beam type of similar capacity and is much simpler in construction. There are no knife edges; it is practically free of all vibrations and noises; it has no moving parts, with the exception of the hydraulic ram, which applies the testing load and also serves to measure the load; it permits the application of a uniform load over the entire range of any test or may be varied within the desired limits; and the load applied at any given moment can be read directly on the dial of a gauge hydraulically connected.

Local conditions will, to a large extent, be a factor in the final decision in choosing one of the two types. That is, consideration must be given to available space, objections to noise and vibrations, and mechanical characteristics. For cramped space, wide range of application, freedom from wear and replacements, quietness, and less expensive machinery, the choice of a hydraulic machine seems to be well supported.

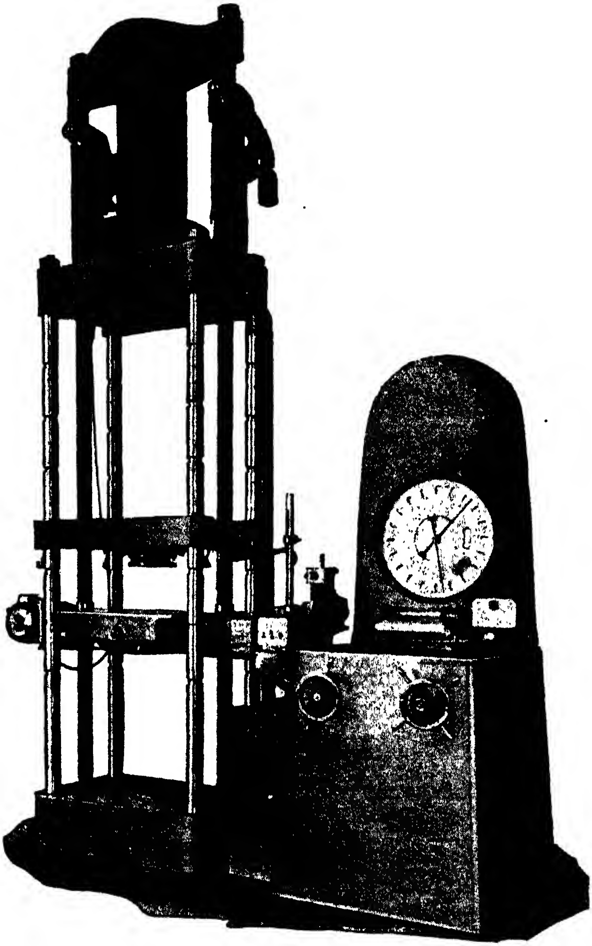

FIG. 2-IV.—Photograph of a hydraulic type testing machine. (*Courtesy of Baldwin-Southwark Corporation.*)

Figure 3-IV shows the A.S.T.M. standard test specimen befor and after testing. It has an original diameter of 0.505 in. hence a cross-sectional area of 0.200 sq. in., and it has tw punch marks 2 in. apart, which constitutes the gauge length Other recommended test specimens for plate, sheet, flat wire strip, and tubing may be found in handbooks where specification will be quoted.[1] Tension specimens of wire and rod are in most cases pulled full size as fabricated. It is well to keep in mind one rule: testing suggests comparison; therefore, in order to get comparison, impose like conditions on all specimens.

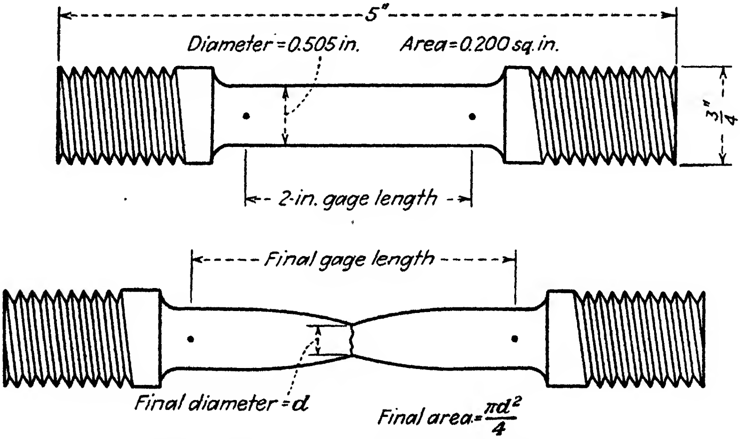

Fig. 3-IV.—Standard tensile specimen before and after testing. (A.S.T.M. Standard.)

Test Procedure.—The experimental details are the same for all types of specimens tested; for example, let us take the standard test bar of soft unhardened steel, clamp it vertically in the jaws of the testing machine, and apply the load at a predetermined rate of speed.

In the first stage of testing, the load will be taken without any visual change in the specimen. The measurement of any elongation or stretching in this period may be accurately measured by the use of an extensometer. Any readings taken will indicate a small constant rate of elongation. Finally, a stage is reached in

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[1] S.A.E., "Handbook," 1943, pp. 354–356; A.S.M., "Metals Handbook," pp. 132–136, Cleveland, Ohio, 1939; Federal Specification for Metals; General Specification for Inspection of, QQ-M-151a.

the testing where, with no drop in the rate of loading, the rate of elongation of the specimen is so rapid that if a beam-type machine is used the beam will drop; if a hydraulic machine with a dial indicator is used, the hand will remain constant. If the extension is being measured by an extensometer, it will indicate a definite increase in the rate of elongation; or, if a pair of dividers is set in the gauge marks at the start of the test, they will tend to jump out. If, just prior to this point, the load should be released, the specimen would return to its original dimensions, but after this point has been passed, a release of the load will show that the specimen has permanently elongated and will not return to its original dimensions.

With further application of load, the specimen will elongate rapidly and, upon reaching a certain value, the specimen will start to reduce in cross-sectional area at a certain point, will neck out, and will finally break.

Now that visual observation has been made of the progressive steps in pulling a test specimen, let us analyze the results and see what experimental data may be obtained from the test that will be useful to the engineer in making and meeting specifications.

Suppose that the load on the test piece is being increased step by step and that we take readings of a series of loads (from the balance beam of the testing machine or from the dial indicator on a hydraulic machine) with corresponding readings of the extensometer until final rupture of the specimen. From these readings there can be plotted a diagram in which the load value is the ordinate and the elongation is the abscissa. Such a diagram is known as a stress-strain diagram.

Stress is the intensity (measured per unit area) of the internally distributed forces or components of force that resist a change in the form of a body. Stress is measured in force per unit area.

Strain is the change per unit of length in a linear dimension of a body—the change that accompanies a stress. Strain is measured in inches per inch of length.

This same diagram may be made autographically. In this arrangement a pen arm plots the load in pounds and a rotating drum plots the deformation in inches. The drum is rotated by means of a cord, which is connected to the moving head of the machine or to an extensometer that clamps on the specimen.

With proper arrangements, this method of plotting becomes very valuable, particularly when it becomes necessary to magnify the curve in order to show the behavior of the material at the point where it is no longer elastic. Autographic recording is best adapted to the hydraulic testing machines.

From the curves (Fig. 4-IV) it is apparent that in the first stage of testing there is a considerable range in which the plot is a sloping straight line. Within this range the material is perfectly elastic, *i.e.*, the stretch is proportional to the load (Hooke's law). The slope in this particular range indicates the value of Young's modulus, *i.e.*, the amount of load or stress for each increment of deformation or strain.

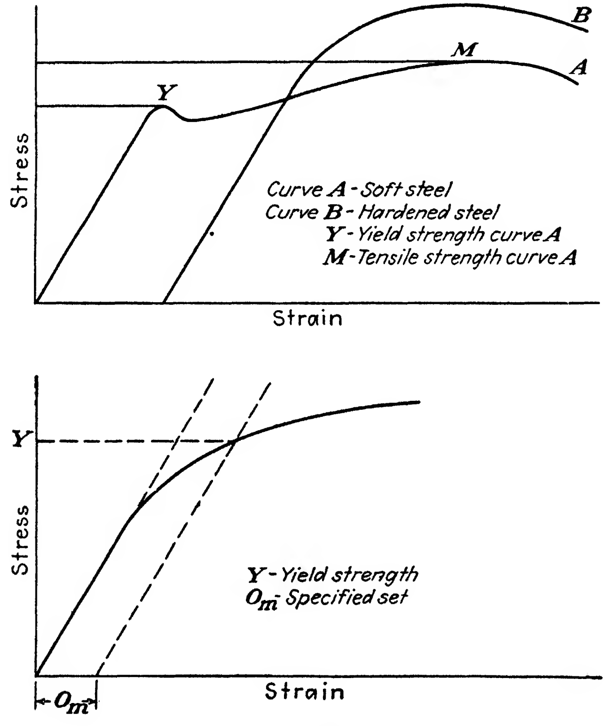

FIG. 4-IV.

The value for Young's modulus, or the "modulus of elasticity," is obtained by dividing the stress (load) by the strain (stretch) where the corresponding values are within the proportional limits. Every metal has its own modulus of elasticity, which is independent of its strength or other mechanical properties and is not affected by heat-treatment or chemical composition. For steel, the modulus of elasticity is approximately 30,000,000 p.s.i., for copper approximately 15,000,000 p.s.i., and for lead about 2,500,000 p.s.i. Practically, this figure may be visualized by considering it as the number of pounds per square inch necessary to stretch the material elastically to double its length. Any metal would, of course, break under these forces but small fractions of this load will function in exactly the same ratio. For example, if a load of 30,000 p.s.i. should be applied to a piece of steel 1 in. in length, it would stretch exactly 0.001 in.

In the previous observations of changes taking place during testing, it was noted that a point was reached where there was a noticeable increase in the elongation with no increase in load. The specimen up to that point had been acting in an elastic fashion; *i.e.*, should the load have been removed, the bar would have returned to its original length. The stress where the curve just deviates from the law of proportionality of stress to strain is known as the *proportional limit.* In earlier studies, the point of deviation from proportionality and the elastic limit, which is the greatest stress that a material is capable of developing without permanent deformation upon complete release of the stress, were separately defined. However, owing to the difficulty in the differentiation and to the fact that the proportional limit is more readily determined, the two figures were considered equivalent and the "proportional elastic limit" was the only one reported. This particular point may be determined graphically by determining the point of tangency of the curve and the straight line, as shown in Fig. 4-IV. Because the location of the point is dependent upon the sensitivity of the measuring device, consistent check by various observers is practically impossible. In addition, it is found that the true proportional limit is lower than the practical load-carrying ability of the material, for under most service conditions the material is capable of withstanding a small permanent deformation. For some materials this figure may be a satisfactory value but the property termed *yield strength,* as determined by the "set method," is found to be a more satisfactory and significant measure of the desired strength property of the material.

The point or stress at which inelastic action starts in the test piece is probably impossible and usually impracticable to determine. For materials that show so-called sharp-kneed stress-strain diagrams, and hence exhibit at a certain stress the special characteristic of yielding without increase in stress, the methods previously described—the drop of the beam, use of dividers, etc.—are satisfactory in most cases for the determination of this point. By these methods the so-called yield point is determined. The yield point is defined[1] as the stress in a material at which there occurs a marked increase in strain without an increase in stress.

The term "yield point" should not be used when referring to material whose stress-strain curve in the region of the yield is a smooth curve of gradual curvature. The determination of the yield point and yield strength would be very difficult in stress-strain diagrams of hardened material, such as cold-worked or heat-treated material, owing to the fact that they exhibit this gradual curvature. The differences in the characteristics of the two types of diagrams is shown in Fig. 4-IV, top.

In order that comparisons of elastic properties of steel may be made, a criterion of a definite amount of permanent elongation has been set up for the yield strength. The yield strength is now defined as the stress at which a material exhibits a definite permanent elongation. The yield strength is, therefore, reported as follows: yield strength (set either 0.1 or 0.2 per cent of the 2-in. gauge length) = 50,000 p.s.i. The determination of this figure is shown in Fig. 4-IV, bottom. The set (0.1 or 0.2 per cent) is laid off on the abscissa (O_m) and a line with that point is drawn parallel with the straight-line portion of the stress-strain line for the material. The point where this line crosses the stress-strain line is the yield strength (Y).

Continuing with the test, as is indicated on the curve, ultimately there is reached the maximum or ultimate stress developed by a given material, the *tensile strength*. Beyond this point "necking" starts effecting a decrease in cross section, the load will drop off, and finally the specimen will break. This final breaking point is known as the *breaking strength*. Tensile

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[1] American Society for Testing Materials, "A.S.T.M. Standards," Part I, Metals (Standard Methods of Tension Testing of Metallic Materials, E 8-42, pp. 898–908), 1942.

strength or ultimate tensile strength is calculated from the maximum load and the original cross-sectional area, and is equal to the maximum load divided by the original cross-sectional area.

After removal of the test from the machine, the bars are placed together and measurement of the elongation after fracture can be made by using a pair of dividers and a scale. This is equal to the ratio of the increase in gauge length following rupture to the original gauge length.

The reduction of the dimensions of the cross section of the test specimen may be measured by a pointed-anvil micrometer. This reduction of area is the ratio of the difference between the original area of the bar and the smallest area after rupture to the original area.

In Table 1-IV are typical data as obtained from a tension test.

TABLE 1-IV.—DIMENSIONS OF TEST SPECIMEN

Dimension	Original state	After testing
Length	2.0 in. (gauge length)	2.2 in.
Diameter	0.505 in.	0.1432 in.
Area	0.2003 sq. in.	0.1615 sq. in.

Test data: Yield point = 18,660 lb.
Ultimate strength = 21,775 lb.

$$\text{Calculations: Yield point} = \frac{18{,}660}{0.2003} = 93{,}160 \text{ p.s.i.}$$

$$\text{Ultimate strength} = \frac{21{,}775}{0.2003} = 108{,}700 \text{ p.s.i.}$$

$$\text{Elongation} = \frac{2.2 - 2.0}{2.0} \times 100 = 10.0\% \text{ in 2 in.}$$

$$\text{Reduction of area} = \frac{0.2003 - 0.1615}{0.2003} \times 100 = 19.4\%$$

OTHER STATIC TESTS

The other type of static tests, those in which there is a steady application of force, are described briefly as follows:

Compression Tests.—This type of test consists in subjecting the material to compressive loads along the axis of the material. The properties usually determined from this type of test are the general compressive-strength properties and the modulus of elasticity of metals.

Shear Tests.—One type of shear test is the torsional, in which the material, which may be either a solid or a tubular specimen, is subjected to twisting forces. The forces in this case are a combination of shear, tension, and compression, all of equal intensity. Frequently, the solid specimens, such as shafts, are tested in full section. The tubular specimens, however, are more suitable for determining the shear-strength values because of the more uniform stress distribution. From this test the modulus of rigidity (yield point in torsion), the modulus of rupture, the ultimate shear strength, and the modulus of elasticity in shear may be calculated.

Other types of shear tests, as for sheet and plate, consist in the use of punches or dies fitted to a testing machine, so that the load required to shear a specimen of a given size can be measured.

Bend Tests.—There are many different types of cold bend tests in use for determining the ductility of different types of metal products. Basically, all the tests consist in bending the material around a die or a mandrel of specified dimensions by the application of a steady force until rupture occurs or until a specified angle has been attained. This type of test is useful for comparing the ductility of materials and testing their stability rather than for determining physical constants. By this means, however, it is possible to determine stresses in tension and compression, modulus of rupture for brittle material, and the modulus of elasticity.

HARDNESS TEST

In considering a hardness test, it is well to define the term "hardness." It is a simple word and one that is used frequently. However, before a definition can be attempted consideration must be given to certain limiting conditions. If a given material is tested by scratching its surface with a file of known hardness, we say that particular surface, if it resists the file, has a file hardness or scratch hardness of surface. Similarly, if this same surface cannot be dented by a diamond or by a hardened ball being pressed against it, or by hitting it with a diamond, we also say that the surface is hard. It so happens, however, that this property of hardness is not the same in the former as in the latter case cited. That is, a steel easily dented may not be scratched and vice versa.

Technically, the term may also mean the resistance to penetration or abrasion. For example, a gear is not hard enough if its teeth are worn down by the rubbing action of the teeth of the other gear with which it meshes. Similarly, a cutting tool is not hard enough if the cutting edge is worn away by contact with the work. These cases are, again, both spoken of in terms of hardness or failure to withstand abrasion.

It seems, therefore, that hardness is a somewhat vague term covering a considerable range of closely related properties. Consequently, it is always best, when referring to hardness, to designate the manner of measurement. Hardness would also seem to fall into two classes: resistance to penetration and resistance to abrasion.

Brinell Hardness.—The Brinell test for the determination of hardness is made by applying a standard load *W*, of 500 or 3,000 kg., to the surface *S*, of the material, through a hardened-steel ball *B*, 10 mm. in diameter. The load, as applied for 10 to 30 sec., results in a not quite spherical impression. The impression would be truly spherical if the ball were not deformed, but since it is slightly deformed and there is a slight recovery of form of the metal, the impression will not be truly spherical. This impression is then measured by a micrometer microscope, the diameter of the impression being converted into the Brinell hardness number by means of the Brinell formula, in which it is assumed that the impression is spherical and is equal to the ratio of the applied load divided by the area of the surface of the impression. The Brinell hardness number is given in kilograms per square millimeter. The formula is as follows:

$$\text{B.H.N.} = \frac{P}{\frac{\pi D}{2}\left(D - \sqrt{D^2 - d^2}\right)}$$

where B.H.N. = Brinell hardness number.
P = load applied.
D = diameter of the ball.
d = diameter of the ball impression.

The Brinell hardness-testing machine is shown in Fig. 5-IV.

The solutions of this formula for many different impressions, using both 500- and 3,000-kg. load, may be found as a table in any of the metallurgical handbooks.

For accurate readings, the following precautions should be observed. Use a well-cleaned surface such as is produced by *grinding, machining, or polishing with emery cloth.* Use a *specimen that is not less than ten times the depth of the impression in thickness.* Do not make the impression so near the edge of the specimen that bulging occurs. Maintain the load 10 sec. for iron and steel and 30 sec. for other materials. Obtain as nearly as possible the true average diameter of the impression. Periodically check the Brinell microscope with a standard glass or scale.

The Brinell test finds its widest practical application on soft and medium-hard materials. On hard materials such as tool steels there is the difficulty of reading the small impressions made. In addition, the ball itself being nothing more than hardened tool steel would tend to flatten against a surface that was of equal hardness. Special balls have been tried to eliminate this difficulty, but none of them has been practical enough to enable the Brinell to compete with the Rockwell and the Scleroscope test on such materials.

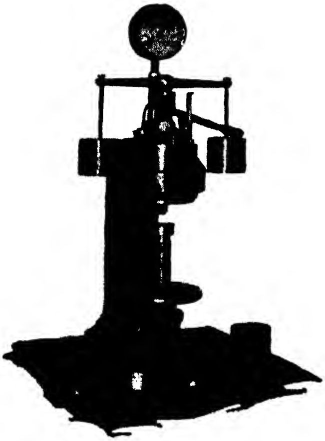

Fig. 5-IV.—Photograph of a Brinell hardness tester. (*Courtesy of Steel City Testing Laboratory.*)

Vickers Hardness.—The Vickers hardness test is an indentation test similar to the Brinell. The test is carried out by impressing into the test material a small diamond pyramid (136°) with a square base just as is done with the ball in the Brinell but the load may be varied from 1 to 120 kg. in varying increments. Since the diamond point is not deformed plastically and only slightly elastically, the hardness values are not appreciably affected by the use of different loads. The depth of the impression can, therefore, be governed by the thickness of the specimen. That fact gives this type of test a much wider application than the ball hardness test. After the impression is made, the diagonals are measured, the load recorded, and the calculations for Vickers hardness are made from the following formula:

$$\text{Hardness} = 1.8544 \frac{\text{load}}{(\text{diagonal})^2}$$

The load is measured in kilograms and the diagonal in millimeters. All the results measured by this test are expressed on a single scale, which is a distinct advantage in the testing of hard and soft materials.

Vickers hardness numbers are found to correspond with the Brinell numbers at low and medium hardnesses but run higher from approximately 300 up. This is shown in Fig. 8-IV. This variation is due to the fact that at high unit pressures the Brinell ball flattens more and more and, as a result, gives lower and lower readings. The diamond pyramid on the Vickers machine is not subject to this distortion.

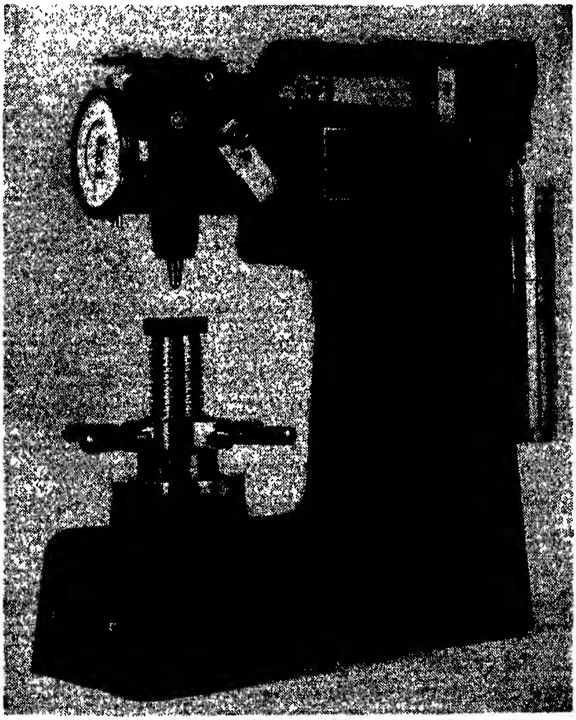

Fig. 6-IV.—Photograph of a Rockwell hardness tester. (*Courtesy of Wilson Mechanical Instrument Company, Inc.*)

Rockwell Hardness.—The Rockwell hardness test consists in the measurement of the depth of an impression of a penetrator into the material. The hardness number is obtained by subtracting the penetration from an arbitrary constant. A minor load of 10 kg. is first applied to the work, which seats the penetrator securely. The dial is then adjusted to the point marked "set" and the lever is tripped applying the major load. After the pointer has come to rest, the major load is removed, leaving the minor load still applied. The Rockwell hardness number may be read directly on the dial at the top of the machine.

In the testing of soft materials a ball $\frac{1}{16}$ in. in diameter is used with a 100-kg. major load. The readings are taken on the B scale of the dial. In the testing of hard and relatively thin materials a 120-deg. diamond-cone penetrator is used with a 150-kg. load. The readings using this penetrator are taken on the C scale of the dial. With the Rockwell C setup the penetration is in the order of 0.005 in., but this is too great a depth for very thin, hard surfaces. In order that these materials may be tested, a Rockwell superficial-hardness tester has been devised, which is similar in characteristics to the usual Rockwell machine, but produces a penetration only of approximately 0.002 in.

The report of each and every test must indicate the scale used, or the readings will be meaningless. The precaution is due to the fact that Rockwell testers can be used with different minor and major loads for work on materials with varying hardnesses, the differently lettered scales being provided to take care of these cases. A Rockwell hardness tester is shown in Fig. 6-IV.

The relationship between the most widely used B and C Rockwell hardnesses is shown in Fig. 8-IV. It is recommended that for correlation the C values be not below 20 and the B values not above 100.

For accurate work the following precautions should be taken, since the accuracy of this machine is dependent upon a measure of the depth of penetration and not on a dead load, as in the Brinell and Vickers tests. Use a flat, clean surface, as curved, scaly, burred, or ridged surfaces give low and false readings. Do not allow the specimen to tilt. Use proper blocks or anvils to fit the specimen. Check the instrument periodically with standard test blocks. Use a specimen of sufficient thickness so that no impression appears on the reverse surface. Mount the machine so that it will be free of vibration.

Scleroscope Hardness.—Scleroscope or Shore hardness is a measure of the height of rebound of a diamond-pointed hammer after striking the test material. The hammer is enclosed in a glass tube, which is provided with a suction bulb whereby the hammer may be raised to the top of the tube and dropped from a definite and fixed height. To make the hardness determination, the instrument (Fig. 7-IV) is held vertical, as indicated by the bob rod on the side, with the bottom resting on the smooth surface of the specimen. The hammer is allowed to drop by compressing the bulb, and the reading, which is the height of the rebound, is noted on the graduated scale.

Erratic results are sometimes obtained with this type of instrument but, with a little practice, assurance of a smooth surface, and solid mounting of specimens, it becomes quite valuable in making hardness tests on specimens too large to be tested by other means.

Monotron Hardness.—Monotron hardness is a measure of the load required to produce a definite penetration. The standard monotron test is made with a diamond-ball impressor, ¾ mm. in diameter, penetrating a standard depth as is indicated by a compensated depth micrometer indicator, of $\frac{9}{5000}$ in. with a given load, measured by a load indicator. The hardness number is then read from the pressure scale in terms of kilograms per unit area with the pressure applied. Upon release of the pressure or load the hand slowly returns toward zero. The extent of the return indicates the amount of permanent deformation, while the remaining figures above zero indicate the elastic recovery factor. With this standard test the scale M-1 is used, which reads directly in Brinell numbers. This test is quite flexible, having the ability to measure the hardness of the softest to the hardest steel, nitrided surfaces, and tungsten carbide, without compensation to the machine or change of impressor points.

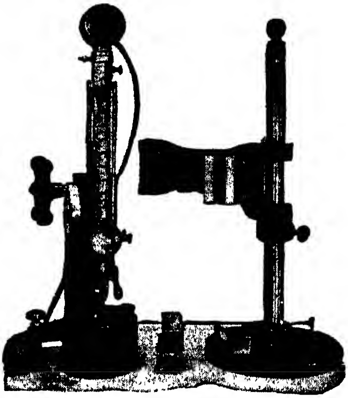

FIG. 7-IV.—Photograph of a Scleroscope hardness tester. (*Courtesy of Shore Instrument & Manufacturing Company.*)

Microcharacter Hardness.—Microcut or Microhardness is a measure of the hardness of microscopic formations or crystals of metals and extremely thin sections. The procedure consists in moving a highly polished and lubricated surface of the material beneath an accurately ground diamond jewel point of three facets, which make a solid right angle, the point of intersection being the cutting point, under a definite pressure of 3 g. for general requirements and 9 g. for harder materials.

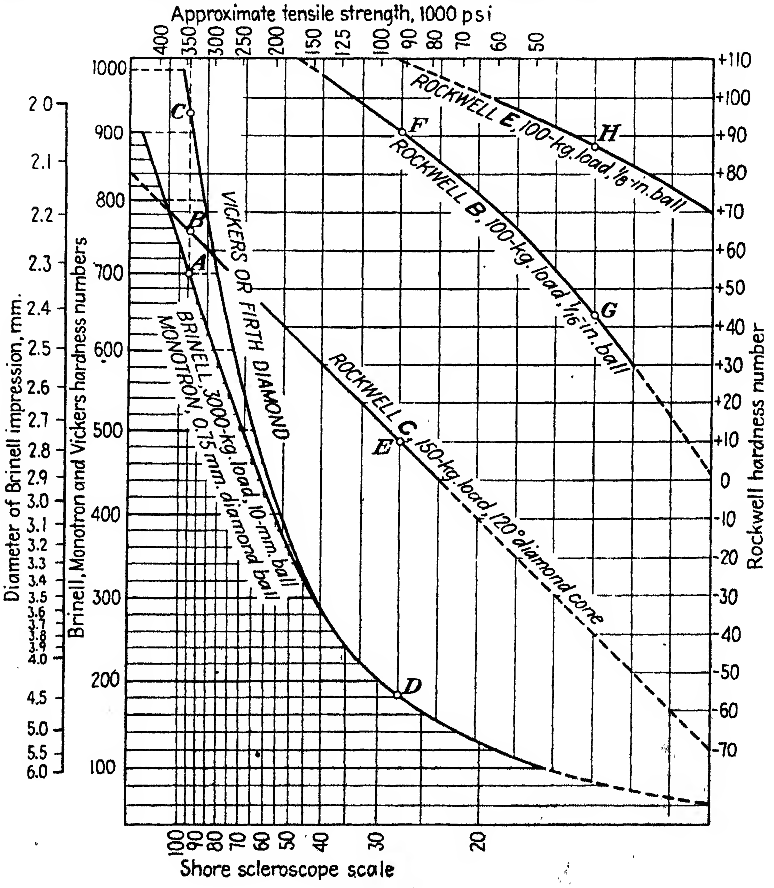

FIG. 8-IV.

In making the measurements, the width is accurately measured by using a highly calibrated optical microscope. Thus it has been determined that the hardness is inversely proportional to the square of the width of the cut. Mathematically, this is expressed by the formula

$$K = \lambda^{-2}10^4$$

in which K represents the microhardness and λ the readings for the width of the cut in microns. The formula is multiplied by 10 and raised to the fourth power in order to avoid long decimals.

Hardness Comparison.—Figure 8-IV[1] shows the general relation between the hardness numbers obtained in the previously discussed tests.

Verticals on this chart represent equivalent hardness. The following examples will illustrate its use:

1. Brinell or Monotron 700 (point A) equals Rockwell C-65 (point B), equals Vickers 930 (point C), equals Scleroscope 89 (bottom scale), and about 340,000 p.s.i. tensile (top scale).
2. Rockwell B-91 (point F) equals C-10 (point E), equals Brinell, Monotron, or Vickers 187, and Scleroscope 29, and has about 92,000 p.s.i. tensile strength.
3. Rockwell B-43 (point G) equals E-87 (point H).

WEAR TEST

Wear, as we think of it, is for all practical purposes a surface phenomenon consisting of the deterioration of the surface or the tearing off of particles of contacting surfaces through friction. The difference in wear of different materials is accounted for by the metals themselves and by the conditions of service. The metals can be more or less fixed but the conditions of service are very complex, so that it is difficult to separate them into components and measure each. Because of these difficulties there cannot be a universal wear test. It is also a definite conclusion that there is no one special metal that is wear resistant under all kinds of conditions.

Among the recognized types or causes of wear are the following:

A. Rolling friction.
   1. Lubricated—as, for example, that between steel rollers and the race in roller bearings.
   2. Unlubricated—as, for example, that between a carwheel and a rail.

B. Sliding friction.
   1. Lubricated—as, for example, that between a shaft and plain bearing.
   2. Unlubricated.

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[1] *Metal Progress*, **24** (No. 4), 723 (1942).

a. Between two solid bodies, such as a brakeshoe and wheel or a wheel and rail, during braking.
b. Between a solid body and an abrasive material of more or less fine grain, as that to which grinding, crushing, and excavating machinery is subjected.

Many types of machines and methods for wear testing have been devised by investigators with the point in view of duplicating conditions of service or developing theories of the builders.

For a detailed description of a number of wear-testing machines and procedures developed to indicate a particular type of wear resistance by some of the pioneer investigators in this field of study, see the reference in the footnote.[1]

NOTCHED-BAR IMPACT TEST

The need for this type of mechanical test, classed as a type of dynamic test, probably arose from the fact that certain materials that proved satisfactory under ordinary tensile tests failed when they were subjected to relatively high impact stresses. Several different types of machines have been designed to measure this figure of toughness, of which the Charpy and the Izod are the most generally used in the United States. The test is used to study the effects of heat-treatment, mechanical treatment, the quality or character of the manufacturing procedure, the presence or absence of undesirable elements, and the physical and chemical homogeneity of the material.

The principle employed in both the Charpy and the Izod machines is the same; both use a swinging pendulum to bring about fracture of a notched specimen. In the Charpy machine the notched bar is placed in a horizontal position on a split anvil, which supports the two ends, and is broken by a horizontal blow opposite the notch by a pendulum or a hammer dropped from a definite height. In the Izod machine the notched bar is held vertical in a vise, with the notch just outside the jaws and facing the direction of the blow. The hammer hits above the notch and breaks the bar by cantilever action. The two methods of setting up the tests are shown in Fig. 9-IV. In both cases the test bar is broken by a single blow and, after fracturing the bar, the hammer swings past the anvil and rises to a height that

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[1] ROSENBERG and HERSCHMAN, Wear of Metals, Bibliography, *Metals and Alloys*, **2**, 52–56 (1931).

depends upon the amount of energy that was consumed in fracturing the bar. The height it will swing is dependent upon the amount of work that was done in the fracture. The work done in fracturing will take a certain amount of energy from the swinging pendulum, with the result that it will not rise to the height it would if its swing had been unimpeded. The energy needed to break the bar is the effective weight of the pendulum times the height of the fall, minus the height of the rise. A scale with a moving pointer gives the readings of the "impact value," either directly in foot-pounds or as the angle of rise, which is converted to foot-pounds.

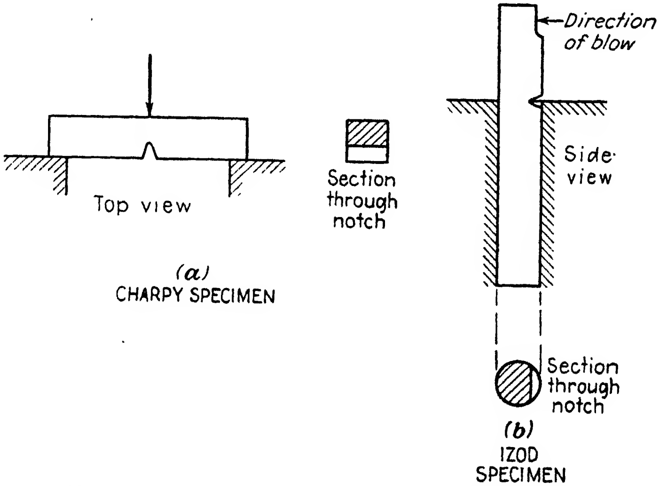
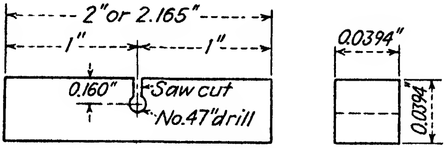


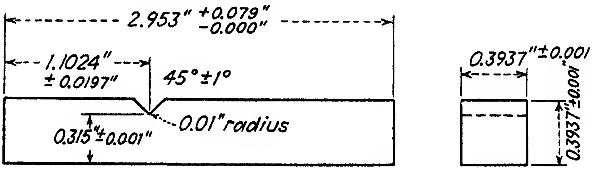
FIG. 9-IV.—Method of loading single-blow, notched-bar impact specimens. (*From Bullens, Batelle, "Steel and Its Treatment," Vol.* I, 1939.)

The types and sizes of the recommended Charpy and Izod test bars may be obtained from any standard handbook. The main difference in the two bars is in the type of notch; the Charpy notch is of the keyhole type, while the Izod is a V-type with a slight radius at the bottom, as shown in Fig. 10-IV. The Charpy bar is always square in cross section so that it can rest on a flat side against the supports, while the Izod bar is usually square like the Charpy but may also be round.

The report of the results of all types of impact test should be made as follows: The force required to rupture or to produce maximum deformation shall be reported in pounds or kilograms; the energy absorbed in rupturing or in producing maximum deformation is to be reported in foot-pounds or kilogram-meters; the velocity of the hammer at the instant of impact, as this factor has some influence on the amount of energy absorbed by some materials; the temperature, as any variation from the usual laboratory temperature of 70°F. causes variations in the results; and the type of notch used.



Charpy notched impact specimen



Izod notched impact specimen

FIG. 10-IV.—Comparison of Charpy and Izod notched impact test specimens.

Factors Affecting Impact Values.—Differences in test-bar dimensions, when using either type of test and any one of the common types of notches bring about certain peculiarities. If test bars of various sizes but proportional dimensions are used, the impact values obtained cannot be readily correlated in any system of units; there is no similarity. Frequently it is found that, even when uniform test pieces from the same steel are used, the values are scattered much more than will be the case with such values as tensile strength or hardness. It is also found that in a series of steels with variable impact, values may range from tough to brittle according to the Izod, in another order by the Charpy with a given notch size, and in another by the Charpy with a different notch size.

The factor of "velocity sensitivity" of steels, *i.e.*, the speed with which the bar is struck, is not particularly significant as far as reproducing results is concerned when the testing is done on impact machines in which the velocities do not vary materially. It is found, however, that the energy absorbed in impact failure at high speeds is usually much lower than that absorbed at moderate speed. In service conditions this may become quite important.

The effect of temperature variation in notched-bar testing is very great. With some steels the variation in laboratory temperatures from summer to winter is sufficient materially to affect the values obtained. In fact, in some cases it is necessary to hold the temperature within ±10°F. in order to obtain consistent results. Such factors as the character of the notch, the analysis, and the heat-treatment all have a very definite effect on the sensitivity of steels to temperature.

The importance of very low-temperature impact tests is apparent when we consider that down to at least −125°F., the strength, elastic properties, and endurance limit steadily increase with little or no change in ductility, but the impact value may drop to a point where it will be no more than 90 per cent of its value at room temperature. Material tested by the usual procedure would fail to show this lack of toughness and, if applied to certain low-temperature service conditions without having the proper tests made, could cause a disastrous failure.

In Table 2-IV is given a set of Charpy notch-toughness tests on a 0.25 carbon steel in the annealed condition at the respective temperatures shown, which will illustrate this effect.

TABLE 2-IV.—TEST DATA

Temperature, °C.	100	50	20	10	0	−20	−30	−40	−50
Results, ft.-lb.	36.1	27.3	19.1	8.8	4.77	3.0	3.7	2.21	2.13

The position from which the sample or samples are taken is also a factor in producing consistent results. If the sample is taken parallel to the direction of rolling, the impact values will be higher than as if it had been taken perpendicular to that direction. The direction of the notch with respect to this directional characteristic will have a similar effect.

Heat-treatment also will have a definite effect. Frequently such treatments may alter the surface, producing decarburization, which tends to raise the impact value, while if the change should be carburization the value will be lowered. Grain size has a *considerable effect on impact values; hence, any unsuitable heating temperatures used in hardening or other treatment that affect* the grain size will have a direct effect on the results. No mechanical test is so sensitive to grain-size changes as is the notch value.

In view of the number of factors affecting the impact test it is not surprising that there is considerable controversy regarding the value of the impact test in evaluating materials for service, *i.e.*, differentiating one material from another. The impact test probably has its chief use in indicating certain variations that may not otherwise be suspected and, in this capacity, indicates that there is a necessity for further study to control them. For a detailed discussion of theory and service relationship of the notched-bar impact test, see the references in the footnote.[1]

Status.—From a commercial standpoint, the conclusions stated at a symposium on impact testing by the American Society for Testing Materials in 1926 may be taken as representing the present-day status. They are as follows:

1. The results are not sufficiently accurate and discriminating to indicate other than large differences in materials.
2. The results are in terms not applicable for use in design.
3. The different types of machines are comparable, but the results of the tests are not.

FATIGUE TEST

Most of us at one time or another have observed failure in springs, axles, shafts, connecting rods, and other moving parts made of metal, caused not by a smashup but by a continued repetition of stresses so low that the effect would be of little consequence if it were not repeated.

Most of us also have heard a garage mechanic describe the cause of a broken axle as being the crystallization of the metal under the continued vibration encountered in the course of service. This has been definitely proved a myth and we now know that this type of fracture is due to a repeated overstressing of the metal, which develops tiny submicroscopic nuclei for

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[1] A.S.M., "Metals Handbook," pp. 671–677, Cleveland, Ohio, 1939; THUM, E. E., Factors Relating the Impact Strength of Metals with Their Service, *Metal Progress*, **32**, 138–141 (1937); AMERICAN SOCIETY FOR TESTING MATERIALS, "Symposium on Impact Testing," **38**, Part II, 21–156 (1938).

failure that in time develop into cracks. This type of failure is known as a *fatigue failure,* the phenomenon of the progressive fracture of a metal by means of a crack that spreads under repeated cycles of stress. Still, when we consider the term "fatigue" of metals, does the metal get "tired" and then rest up? Endurance or progressive fracture, the resistance to progressive failure under repeated stresses, far better describes the condition.

In referring to stress values, it becomes important that the designer should know the actual stress that will be imposed upon a certain material under given service conditions. Frequently the designer will estimate the stress as being low, when, as a matter of fact, he has failed to account for the actual stress values, for example, those which are actually quite above the calculated value, at an abrupt change of section, at a sharp groove or slot, or at some internal or external flaw.

The testing engineer must, therefore, design his machine and his specimens so that the actual stress is the calculated stress. The engineer can then determine the endurance limit or a limiting stress below which the metal will withstand without fracture an indefinitely large number of cycles of stress. The designer from these results must endeavor to fix his design so that the part will never be subjected to stresses above the endurance limit. One of the weak features of this type of test is shown here. This standard test demonstrates what the metal will stand under favorable conditions; but it fails to show what abuse the metal will withstand, how often it may accidentally be taken above the endurance limit without failure occurring. In such a case, where the stresses cannot be controlled, the designer uses a large factor of safety, *i.e.,* the material must be chosen so that it will not be severely damaged by overstress.

Classes of Testing Machines.—Testing machines for the determination of the endurance limit of metals may be classified according to the type of stress that they are called upon to duplicate. This classification is as follows: (1) direct-tension compression, (2) flexure or bending, (3) torsion or shear. This entire classification is subdivided with respect to the methods of development of the stress, as (*a*) constant-stress machines, in which the stress is constant throughout the test, and (*b*) constant-strain machines, in which the deformation is kept constant throughout the test.

Direct-tension Compression.—The direct-tension compression is practically self-explanatory; it involves merely the repeated application of alternate compressions and tensions.

Flexure and Bending.—The flexure and bending type of test is the most widely used, being applied in more than 90 per cent of the tests made. This type may be further subdivided with regard to two methods: (1) the simple rotating-beam type, which is further divided into two modifications, the long- and short-specimen type; and (2) the cantilever rotating-beam type.

The rotating-beam type of test in general is made on a cylindrical test piece supported as a beam, with loading at two points. The specimen is rotated so that there is an alternate shift of the uniform bending moment between the load points from tension to compression every 180 deg. of rotation. The difference between the long- and short-specimen type lies in the length of the specimen. The short-type test specimen is of only sufficient length to extend between the load points. This short type of test offers certain advantages from this viewpoint, that the moment arm is constant regardless of the size or shape of the specimen, while the moment arm in the long type may be varied with respect to the specimen and thus introduce considerable possibility of error.

It is of particular interest to note here that in cylindrical specimens the stress is not uniformly distributed over the entire cross section. The material that is being stressed to the maximum and being used in the determination of the endurance limit varies from maximum at the outer fibers to zero at the axis. In other words, the fibers at the top of the specimen are in compression and the fibers at the bottom are in tension.

The cantilever rotating-beam type of test requires an overhanging beam of the cantilever type with a weight hanging from the end of the specimen. In this test, if the specimen is of uniform cross section, the bending moment varies from zero at the load point to maximum at the support. The stresses vary as the bending moment and alternate with respect to compression and tension for every cycle.

The reverse plain bending type of machine is used for flat stock and sheet metal and performs a reverse bending operation. In most cases the machine produces a constant strain in which the stress in the specimen is computed from the deflection.

Torsion or Shear.—The torsion testing machine develops a condition of shearing stress, which is alternately reversed in direction. The stresses are uniformly distributed over the entire specimen but vary from maximum at the outer fibers to zero at the center of the cross section.

In order to determine the endurance limit, it is necessary to prepare a number of similar specimens, all representative of the material. The first specimen is tested at a relatively high stress, so that with a small number of stress applications failure will occur. The other tests are then made, a lower stress being employed each time. The number of applications of stress will increase as the applied stress decreases until a stress is reached where the specimen will have an exceedingly long life.

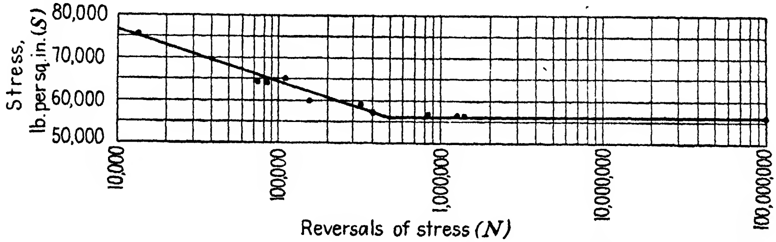

FIG. 11-IV.—Typical S-N or stress-cycle diagram for a fatigue test. Semi-logarithmic plotting.

If these tests are then plotted on either logarithmic or semi-logarithmic paper with the values of stress (S) as the ordinate and the number of repetitions to cause failure (N) as abscissas, the result is an SN diagram (Fig. 11-IV).[1] The type of diagram shown here is a typical one, in which we have two straight lines, one sloping and one horizontal, joined by a slightly curved "knee." The horizontal line indicates that a stress value has been reached that will require an enormous number of repetitions of stress to produce failure. The last points show that the specimen has not failed at 100,000,000 repetitions of stress.

The value of the endurance limit is determined by running sufficient tests so as to locate the "knee" of the endurance curve. In this particular curve, $N = 1{,}000{,}000$. On some steels, N will not be reached until it is 10,000,000 but N has never exceeded 10,000,000 for any ferrous material.

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[1] TEMPLIN, R. L., Fatigue Testing, A.S.M. "Metals Handbook," p. 145, Cleveland, Ohio, 1939.

The straight endurance limits of steels under the best of conditions are found to be pretty much proportional to the tensile strength or Brinell hardness. The ratio, for example, of endurance limit to tensile strength is around 50:100. However, these ratios are such that the actual test can hardly be dispensed with, because, if only service and no abuse were to be considered, the higher the tensile strength, the higher would be the endurance limit. This ratio is further affected and does not hold true for quenched and heat-treated steels in which the internal stresses have not been released. Either the relationship of endurance limit with other physical properties has shown no correlation, or investigations have not been carried far enough to make a definite statement regarding comparative properties.

The endurance limit gives only one kind of information and does not specify that its results, when applied to service conditions, are the best from an endurance test alone. For a more complete study of such conditions as low temperature, high temperature, and corrosion on fatigue, see the reference in the footnote.[1]

CORROSION

It is seldom that one becomes confronted with the problem of corrosion or, rather, seriously considers it, as it never appears as a separate item on the ledger. This problem, however, is of major importance and interest to both the producer and the consumer of iron and steel products.

Further to emphasize its seriousness and importance, let us consider the principal agencies of destruction at work on metals. They are corrosion, wear and erosion, high and low temperatures and temperature fluctuations, and repetitive stresses and accidental overpowering impacts. Of these, corrosion is the most destructive, its rate varying with the environment on unprotected iron and steel products from an insignificant amount to ½-in. penetration per year. From a monetary standpoint it has been estimated that the world cost of replacement and prevention is approximately two billion dollars annually.

For a fuller conception of corrosion's importance and the enormity of its effect on everyone, we have only to consider the great number of industries affected. Space will not permit a

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[1] A.S.M. "Metals Handbook," pp. 147–153, Cleveland, Ohio, 1939.

discussion of the results in each industry, but pause to think how corrosion affects the following industries and how it affects you through them—paper mills and the petroleum, chemical, pharmaceutical, dairy, brewing, smelting, architectural, and automotive concerns.

Corrosion Theories.—During the past 50 years several theories of corrosion have been advanced, but all others have been discarded in favor of the one now accepted by the majority of investigators, the electrolytic theory.

Corrosion was at first thought to be a case of simple oxidation, or the reversion to the natural state; *i.e.*, iron oxide, as in the case of heating iron in air. Iron oxide was further believed to consist of a combination of hydrates, oxides, and carbonates, mostly carbonates. This viewpoint led to the acid theory of corrosion, which states that the presence of an acid is necessary for corrosion to start. Therefore, in the case of corrosion in the atmosphere or by immersion, the main acid present was carbonic acid (H_2CO_3). This led to many investigations regarding the relationship between CO_2 and corrosion. Among the more notable ones were those by Cushman and Gardner, and by Whitney, who proved that CO_2 was not necessary for corrosion. Whitney's experiment in 1903 led to the electrolytic theory of corrosion, which covers substantially all the cases of corrosion, in which the driving force of the corrosive reaction between the metal and environment is electrochemical.

The electrolytic theory involves the fact that every metal when placed in water is subjected to a fixed tendency to go into solution until the water becomes saturated with the metal ions and equilibrium is established. This involves water free of oxygen and an acid radical. In the case of iron, iron will go into solution and first form $Fe(OH)_2$

$$Fe^{xx} + 2HOH = Fe(OH)_2 + H + H \qquad (1)$$

ferrous hydrate, which, owing to its solubility, will not be noticed until oxygen has been admitted to the solution, when a further reaction takes place with the formation of $Fe(OH)_3$, ferric hydroxide, which settles out of the solution as a reddish-brown precipitate and passes to Fe_2O_3, or rust. During the first reaction involving $Fe(OH)_3$, hydrogen will come out of solution, plate on the metal, and form a polarizing film, which will retard or stop the corrosion. Corrosion may then proceed only by the removal of this hydrogen film. This removal is brought about in two ways, according to the following reactions:

$$H + H = H_2 \tag{2}$$
$$H + H + O = H_2O \tag{3}$$

Therefore, in the case of corrosion of iron in neutral or basic solutions, it becomes necessary that oxygen shall reach the metallic surface. This reaction involves a secondary reaction involving the oxidation of $Fe(OH)_2$ to $Fe(OH)_3$ but the most important consideration is the fact that, as the alkalinity of the solution decreases, the number of hydrogen ions increases until a point is reached where reaction (1) becomes the determining factor. The rate of corrosion will be, therefore, inhibited but not necessarily stopped.

In the case of iron in an acid solution, reaction (2) becomes the controlling factor as soon as there is an evolution of hydrogen. Corrosion may also be maintained in the absence of oxygen with the evolution of hydrogen gas or, in other cases, by direct chemical attack.

The principal factors[1] involved in corrosion are as follows:

1. Factors with respect to the metal:
    - a. Chemical reactivity of the metal.
    - b. Chemical and physical homogeneity and texture of the metal surface.
    - c. Inherent ability of the metal to form an insoluble protective film.
    - d. The overvoltage of hydrogen; *i.e.*, the difference between the potential required for the passage of the current and the equilibrium value.
    - e. Internal stresses.
2. Factors external to the metal, which vary with the environments:
    - a. Hydrogen ion concentration.
    - b. Effective supply of oxygen in the solution adjacent to the metal.
    - c. Distribution of the available oxygen on the metal surface.

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[1] Speller, F. N., The Corrosion Problem with Respect to Iron and Steel, *Trans. A.I.M.M.E.*, **113**, 13–33 (1934).

d. Nature, concentration, and distribution of other ions in the solution.

e. Effective rate of flow of the liquid past the metal surface.

f. Presence of solid particles or a coating of any kind on the metal surface or contact with other conducting material in the presence of an electrolyte.

g. Temperature.

h. Static or alternating cycles of stress (corrosion fatigue).

i. Ability of environment to build up protective films on the metal.

Preventive Measures.—To a certain degree, corrosion may be prevented by one of the following means:

1. Changing the environment; for example, in positions where the medium is limited, it may be made less active by the addition of a neutralizer or a passivator, or by the removal of the oxygen.

2. Isolation of the metal from the corrosive environment, which includes all types of protective coatings, metallic and nonmetallic, of varying thicknesses.

3. Building up the resistance of the metal to corrosion, particularly with reference to certain corrosion-resistant alloys. It is interesting to note that only such elements as chromium, nickel, copper, silicon, molybdenum, and aluminum have been found to increase the corrosive resistance of iron.

Corrosion Tests.—Laboratory tests for determining the corrosive deterioration of iron and steel are usually short-time or accelerated tests. In other words, they attempt to predict what may be expected after years of exposure to service conditions. The laboratory tests usually made are the salt-spray, variable-weather, intermittent-immersion, total-immersion, and chemical-corrosion tests. The service tests usually made are the underground, chemical-corrosion, and atmospheric-corrosion tests.

It must further be kept in mind that, because of the many factors involved, there is no all-round test for corrosion. However, an approximate indication may be had as to the suitability of certain alloys to resist certain types of corrosion and it may be had only by using tests adapted to each particular type.

The interpretation of the results may be made in a number of ways, but they should always be based on the relationship of the results to service conditions.

The general procedure is to note the appearance of the corroded specimens, the adherence of the products of corrosion, and the distribution of the corrosion over the specimen.

The corrosion products should be removed and the extent of the damage determined by one of the following three methods:

1. The loss in weight of the specimen—a unit that is expressed in average inches penetration per given length of time and may be plotted as a time-rate curve.
2. The maximum depth of penetration—a unit that is usually expressed in penetration for a given length of time. A useful practical indication may be had here of the uniformity of attack in comparison with the loss in weight (1).
3. The subsurface attack may be measured by determining certain physical properties of the metal before and after corrosion. A microscopic examination is also advisable in this case.

Although we have at our command certain established facts concerning corrosion, it should always be kept in mind that in all cases laboratory results should not be relied upon unless there has been a previous correlation with service conditions or tests; *i.e.*, the testing environment should be truly representative.

CREEP TEST

Previous to 1920 such industries as the power, petroleum, and chemical were operating their units at such temperatures and pressures that the conventional materials of construction were quite suitable. Since that time, however, the need for more economical operations has driven the designer to make use of increasing pressures and temperatures in his units. It was ultimately found that at elevated pressures and temperatures, ordinary materials had distinctly different physical characteristics. It was found that construction metals of drums, tubes, autoclaves, and fittings have a tendency at an elevated temperature to "creep," or slowly stretch under stress.

In the years prior to 1920, physical testing of metals was conducted almost exclusively at atmospheric temperatures by the short-time tensile test. The properties thus determined, with a proper factor of safety to take care of unknown variables, were assumed by the designer to be safe for equipment operating at moderate temperatures. With the advent of higher pressures and temperatures, the physical properties, as determined by the short-time tensile test, no longer yielded the proper information. This study has, therefore, resolved itself into a determination of the ability of a metal to undergo continuous deformation at an elevated temperature—even to final rupture—under prolonged loading at stresses considerably below its strength as determined by the ordinary tensile test at that temperature. This property has become known as the *creep* or plastic *flow* of the metal.

The measurement of the high-temperature properties of metals was first attempted on the ordinary tensile machine by heating the specimen to the required temperature, transferring it quickly to the testing machine, and breaking it by the usual procedure. As the test is used today, the tensile machine is fitted with an electric-tube furnace to heat the specimen, some sensitive type of extensometer is used to measure the elongation of the specimen very accurately, and a definite and controlled rate of stress application is maintained.

The proportional limit as determined by this test was formerly thought to be indicative of the load-carrying ability of the metal at the temperature of the test. It was soon found, however, that at high temperatures the proportional limit became very uncertain and, furthermore, was dependent upon the rate of loading. The strength of the metal, as determined by the test, also decreased as the rate of loading was decreased, very slow application of loading requiring a breaking stress far below that needed at the normal rate of application usually used for atmospheric tests. The present opinion of most metallurgists is that at temperatures up to 550°F., using the refinements indicated above, the proportional limit of steels as determined by carefully conducted short-time tests is valid for design purposes. Above that temperature, the uncertainty in the determination of the exact point of divergence from Hooke's law and the effect of rate of loading cause the results to be of doubtful value.

The realization of the above defects in the short-time tensile test led to the evolution and use of the long-time, or creep, test. The aim of such a test is to determine the relation between temperature and stress that will produce deformation at some fixed rate. This stress is called the *creep strength* and is expressed by the engineer in terms of the stress required for a rate of creep of 1 per cent creep per 100,000 hr., 1 per cent per 10,000 hr., or 1 per cent per 1,000 hr. Obviously, a testing program designed to investigate the properties of any particular material for a particular class of work over a period of say 15 years is impractical. Since few, if any, laboratory creep tests have been run over 5,000 hr. and since the exact determination of a final creep rate of 1 per cent in 100,000 hr. would require more than the usual care, test figures of such phrasing are the result of extrapolation of the data.

Testing Equipment.—In general detail, the testing equipment that is needed for this type of work consists of the following parts:

1. A heavy steel frame to support the apparatus rigidly in place and to avoid undue vibration of the lighter parts.
2. A set of jaws to hold the specimen, the upper one being connected to the frame and the lower to a lever, by means of which the load is applied to the specimen.
3. An electric-tube furnace, large enough to accommodate the specimen and equipped with automatic temperature control.
4. Some means of measuring the extension of the specimen. This can be done either by a dial type or by a mirror or mirror-prism optical-type extensometer, or the extension can be measured directly by observing the distance between two fixed gauge marks with a traveling telescope.
5. The specimen is equipped with a thermocouple to measure its temperature. The specimen is usually of the standard tensile-test-piece design, except that the gauge length is sometimes longer than the standard 2 in.

Testing Procedure.—The following four methods of testing are in general use:

1. The single-step method consists in the application to the specimen of a definite and fixed load at a constant temperature until the extension becomes zero or remains constant over a period of several hundred hours. The method has the advantage of most nearly paralleling service conditions and its results are believed to be the most reliable for use in high-temperature design. Its disadvantage, the fact that a separate piece of equipment is necessary for each temperature and stress, has led to the development of the following tests to get away from this disadvantage.

2. The step-up method consists in subjecting a specimen, at constant temperature, to a small stress until the rate of flow is constant from 200 to 400 hr. The stress is increased in increments and the procedure repeated.

3. The step-down method consists in subjecting a specimen, at constant temperature, to a high stress until the rate of flow is constant. The stress is lowered in increments and the procedure repeated.

4. The step-temperature methods consist in subjecting a specimen, at constant stress, to a given temperature until the rate of flow is constant. The temperature is raised in increments and the procedure repeated.

Let us now consider a typical time-elongation curve from a test as is shown in Fig. 12-IV.[1] This curve gives the results on a low carbon steel with stresses of 9,225 p.s.i. and 7,500 p.s.i., respectively. The result of the unit stress of 9,225 lb. shows the development of three stages of creep: (1) an initial flow, (2) a secondary flow at a lower rate than that observed in the first period, and (3) a final, rapid flow to fracture.

Many curves have only the first and second stage, as is shown in the case of the 7,500-lb. loading, in which the flow is occurring at a small, but uniform, rate with a possible slight decreasing tendency, even though the test is carried out for 2,000 hr.

The conception of creep[2] that appears to be most satisfactory, consistent with the observed behavior of metals subjected to stress at elevated temperatures, may be summarized as follows: At elevated temperatures at or below the equicohesive temperature, which is the lowest temperature of recrystallization, the deformation or creep resulting from the application of a given stress over a given time period is a running balance between the yielding of the material and the strain hardening caused by such yielding. At or below this range, strain hardening tends to predominate and continuous measurable creep will not occur unless the stresses are sufficiently great to overcome the resistance caused by strain hardening. In fact, the yielding of steel at room temperature when the proportional limit is exceeded is in reality

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[1] WHITE and CLARK, Creep Characteristics of Metals at Elevated Temperatures, *Trans. A.S.S.T.*, **21**, 5 (1933).

[2] CLARK and WHITE, Creep Characteristics of Metals, *Trans. A.S.M.*, **24**, 831–869 (1936).

a creep phenomenon, but unless the stress is sufficiently great the measurable deformation soon stops as a result of the marked rate of strain hardening.

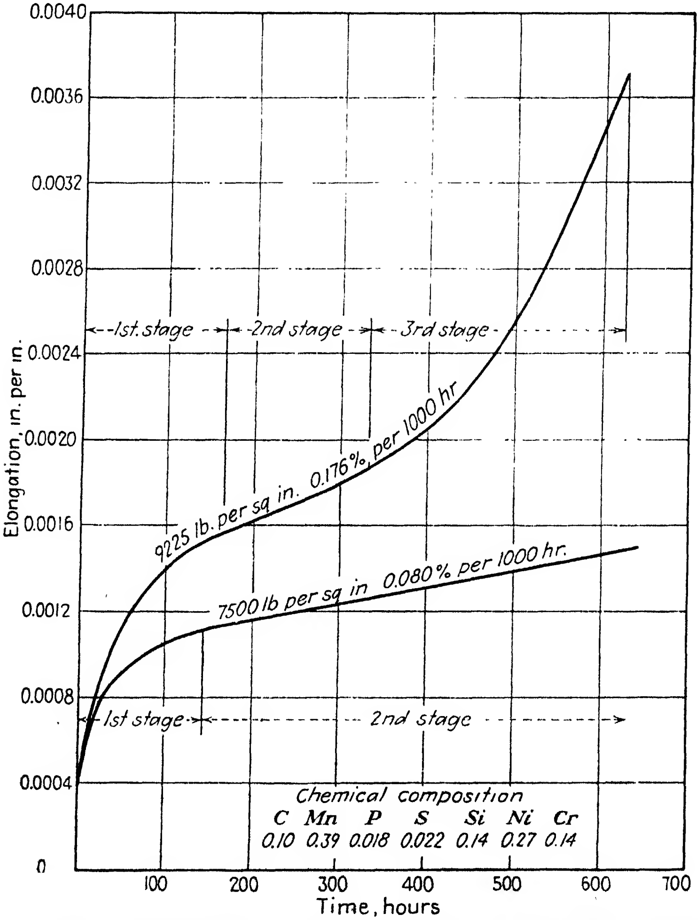

FIG. 12-IV.—Time-elongation curves for carbon steels at 1000°F. (540°C.).

At temperatures above the equicohesive temperature, however, the recrystallization rate exceeds the strain-hardening rate, and creep will proceed under the lowest stresses. In fact, it is believed that creep could be detected at these higher temperatures under any stress whatsoever if sufficiently sensitive apparatus were available.

It is also believed that the mechanism of deformation changes as the testing temperature is increased. At temperatures below the equicohesive temperature, deformation apparently occurs largely within the crystals themselves and, since these crystals are perfect elastic bodies, the metal exhibits elastic properties. At temperatures above the equicohesive temperature, however, it is felt that if sufficient time is allowed the deformation occurs through the movement of the crystals themselves, *i.e.*, within the materials surrounding the grains. Regardless of the nature of this material and whether it is amorphous or severely strained crystals, there is general agreement that it is not elastic. Accordingly, metals above this critical temperature exhibit plastic rather than elastic properties.

Factors That Influence Creep Characteristics.—The creep resistance of a material appears to be a rather sensitive property and it is influenced to a greater degree by changes in the material itself or in the testing methods than are the more common physical properties. Among the factors that must be considered as having a possible influence on the high-temperature creep characteristics of metallic materials are the following:[1]

1. *The Melting Process.*—Noticeable differences are sometimes found in the behavior of open-hearth, electric-arc, and induction-furnace steel. The differences are not always marked, and too-broad generalizations are not always permissible because of the presence of other variables.

2. *Melting Practice.*—The practice employed in making steel may greatly influence its creep characteristics. Rimmed carbon steels possess lower creep strength than do killed steels, especially at the more elevated temperatures. Even in killed steels, however, the characteristics may vary considerably, depending on the type of deoxidation.

3. *Chemical Composition.*—The creep strength is greatly dependent upon, but not established by, the chemical composition alone. It is indicated that both the type and the amount of alloying element added are of importance.

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[1] AMERICAN SOCIETY OF MECHANICAL ENGINEERS and AMERICAN SOCIETY FOR TESTING MATERIALS, Compilation of Available High Temperature Creep Characteristics of Metals and Alloys by Creep Data Section of A.S.M.E.—A.S.T.M. Joint Research Committee on the Effect of Temperature on the Properties of Metals, 1938.

4. *Grain Size.*—It is found that, at temperatures below the lowest temperature of recrystallization, fine-grain steels have the greatest creep resistance, while, at temperatures above, coarse-grain steels are superior.

5. *Heat-treatment.*—Heat-treatment has been definitely found to influence the resulting creep resistance and other high-temperature properties. The most suitable treatment varies, however, depending upon the analysis, the proposed operating temperature, and the desired room-temperature physical properties. In general, but not always, a spheroidizing treatment lowers the creep strength.

While it is important that the creep strength should be known and that it should be sufficient to withstand the desired stresses, we must keep in mind that the creep test is only one factor in the appraisal for high-temperature service. Even though the creep values obtained are satisfactory, there still may be some changes not discernible during testing that may make the material of little value. Therefore, even though the material should be well within the limits of specification, further examination should be made of the specimens, following creep testing, for such matters as signs of lack of structural stability, *e.g.*, changes in hardness, magnetic properties, microstructure, resistance to impact, and resistance to oxidation and corrosion. This will give a warning of any harmful effects that have begun during the test and will show whether the data may be taken by the designer at their face value.

For a more complete study reference should be made to the authority in the footnote.[1]

Suggested Questions and Problems for Study and Class Discussion

1. How might the testing engineers of an automobile concern apply every test we have discussed in properly designing their motor cars?

2. Are these tests being applied in your metallurgical department? How?

3. Explain what procedure and what tests you would apply in investigating the following problems (obviously we shall have to neglect microstructure):

- *a.* A certain oil company is intending to build a pipe line from Philadelphia to Pittsburgh for the purpose of transporting gasoline. Long service life and low cost being prime objectives, they investigated welded and seamless copper and noncopper bearing open-hearth pipe, along with wrought- and cast-iron pipe.

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[1] Kanter, J. J., The Creep of Metals and Alloys, A.S.M. "Metals Handbook," pp. 504–511, Cleveland, Ohio, 1939.

b. A certain oil company in rotary drilling four wells, experienced an abnormally high number of twist-offs with several strings of seamless drill pipe. For identification purposes this company stamped a number on the body of the pipe quite close to the upset. The strings were made from the same heat of steel, which was quite normal, with the submitted sections checking upon being analyzed.

c. A certain automobile concern experienced considerable difficulty with breakage of the wire being used as spokes.

d. A certain steel company was submitting a trial order of cold-rolled tin plate to a certain manufacturer for a deep-drawing job. The steel company processed plate made from three types of steel, namely, rimmed, aluminum-killed, and a special practice. All the plate was given the same processing.

e. A certain automobile manufacturer was encountering considerably higher breakage than was normally allowed, in making the difficult draw of the front fender on their car. On this particular job the required steel analysis was C, 0.08; Mn, .20/.35; P, 0.04 max.; S, 0.05 max.; Si, 0.02 max.; and the size was 0.039 by 62 by 85 in. No unusual variances occurred during processing.

f. Determine the ultimate strength, yield point in pounds per square inch, the elongation (in 2 in.) and reduction of area in percentage from the following data:

Dimensions of Test Bar

Dimension	Before pulling, in.	After pulling, in.
Gauge length	2	2.55
Diameter	0.505	0.160

Machine readings:

Yield point = 9,120 lb.
Ultimate strength = 17,500 lb.

g. From the figures in the accompanying table, what would be your opinion on the correlation of test data between different types of impact testing machines?

Type	Test data	
	At room temperature	At −75°F.
Charpy impact	190	5
Izod	90	90
Charpy	55	30

h. By test the endurance limit of a given material has been determined to be 60,000 p.s.i. What interpretation can you make for this figure?

i. Explain the meaning of the term "creep." Why are creep tests conducted?

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CHAPTER V

CONSTITUTION OF METALLIC ALLOYS

ELEMENTARY BINARY DIAGRAMS

The study of the phenomena that occur during the heating and cooling of metallic bodies has given to the scientist a new understanding of the nature of metals and alloys. These phenomena, which relate to fusion and solidification and to many internal changes that occur in metals or alloys while they are partially or wholly solid, constitute an extremely important branch of physical metallurgy. The importance of this phase lies in the fact that our most comprehensive and satisfactory method of describing the nature and constitution of a given system consists of a diagram, known as an *equilibrium diagram* or a *constitutional diagram,* which is based on thermal data. This diagram, besides being representative of the behavior of the metals and alloys at different temperatures so far as solidification and fusion are concerned, really relates to all the physical properties of the materials. This connection is, therefore, useful for the purpose of predicting properties of a given alloy of a given composition from its position on the diagram.

The constitution of metals and alloys is revealed by two means: (1) through the mechanism of their solidification, as disclosed by their equilibrium diagrams; and (2) through a microscopic examination of their structure after solidification.

Solidification of Pure Metals.—Let us first study, by means of a pyrometer, the thermal conditions that prevail during the cooling of a pure metal, from temperatures where it is molten to ordinary room temperatures. Suppose that, during the cooling, simultaneous readings of time and temperature are taken and that these are plotted as coordinates on a diagram. The resulting curve, known as a *cooling* curve for the metal, will be very similar to that given in Fig. 1-V. At point *A* on the figure, the metal is molten, its temperature being denoted by *T*. The cooling from point *A* to *B*, with the temperature falling from T to T_s, is uniformly retarded, as is shown by the gradually decreasing slope of the curve. The curve indicates, upon reaching temperature T_s, the occurrence of a critical point, which corresponds to the horizontal portion of the curve BC. It is apparent that, at this temperature, the fall is arrested for a time interval represented by $t - t'$. On continued cooling the normal rate is continued until we reach room temperature, as is indicated by CD, which, were it not for the horizontal portion BC, would be a direct continuation of AB. This point of arrest immediately becomes associated with the freezing point of the metal.

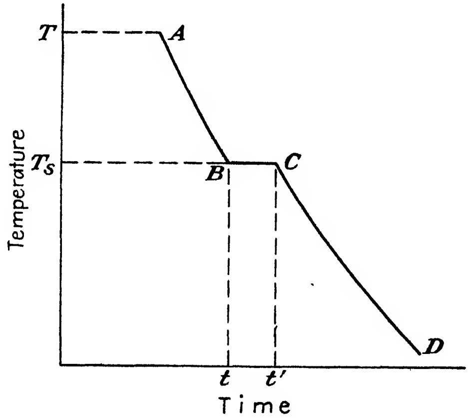


FIG. 1-V.—Typical cooling curve of pure metal.

Although the metal was exposed to the same conditions both above and below this point, its temperature remained constant, indicating that the heat liberated must be equal to the heat lost through radiation and conduction. The heat thus evolved is known as the *latent heat of solidification*. Therefore, in the figure, AB represents the cooling of the liquid metal, BC its solidification, and CD its cooling after solidification. If this same metal were to be heated from room temperature to above its melting point, a similar curve would be obtained under normal conditions. This constancy of solidification, as exhibited, is typical of only three classes of bodies: elements, pure compounds of two or more metals, and a special class of alloys known as *eutectic* alloys. The two last named classes will be discussed fully later in the text.

The typical structure of a pure metal is illustrated in Fig. 2-V, which shows pure iron at 100 diameters. This structure is said to consist of polyhedral grains or crystals of iron. The usual terminology designates the shapes as grains and the total mass as the grain structure of the metal. Technically speaking, the terms "grain" and "crystal" are interchangeable in this case, since the shapes that have been arbitrarily termed *grains* are also crystals, in that within each shape there exists a definite relationship between the atomic layers of materials, which is referred to as its lattice structure.

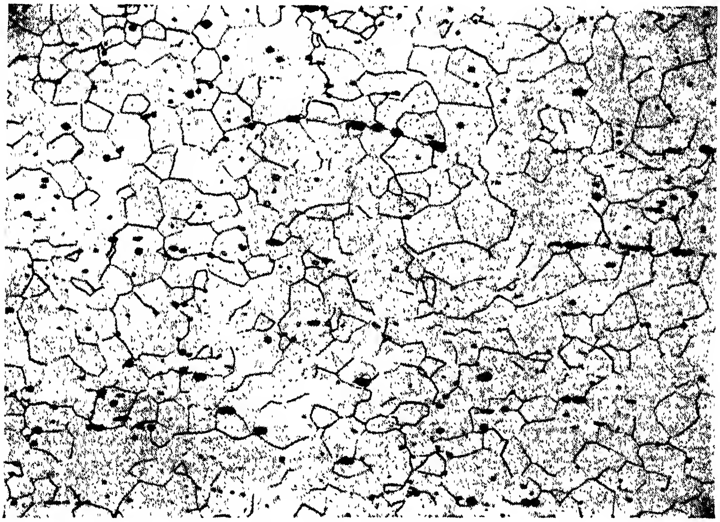


Fig. 2-V.—Commercially pure iron.

This may be further illustrated in the case of quartz (SiO_2). The crystal structure of quartz has a very definite crystallographic form, where the atoms and the molecules have a definite relationship to each other. If this crystal is fractured, as it will be in sand, the resulting material is referred to as grains.

Solidification of a Solid Solution.—Now suppose that a small quantity of a second metal that is soluble in the first is added, that the mixture is melted as before and allowed to cool, and that a cooling curve is plotted. A number of so-called *binary* alloys yield a cooling curve of the type shown in Fig. 3-V. As the alloy cools from T to T_b or from A to B the curve is smooth and, therefore, denotes a normal rate of cooling. At *B*, however, there is a sudden shift in direction, so that from temperature T_b to T_c the cooling is slow. From T_c to room temperature the cooling again becomes normal. Since *BC* on the curve indicates that there must be an evolution of heat, resulting in a marked retardation in the cooling for a time lasting $t - t'$ sec., it may be inferred that this interval corresponds to its solidification. Therefore, binary alloys that yield such cooling curves are unlike pure metals, which solidify at constant temperature, in that their solidification lasts for a time $t - t'$ sec. while the temperature drops from T_b to T_c. In the figure, therefore, *AB* indicates the normal cooling of the molten metal, *B* where solidification starts and *C* where it is complete, $t - t'$ the time interval required for solidification, $T_b - T_c$ the temperature drop during solidification, and *CD* normal cooling of the solidified alloy. At point *B* the alloy is liquid and is accordingly called the *liquidus* point. At point *C* the alloy is solid and is called the *solidus* point. Above point *B* the alloy is liquid and below point *C* the alloy is solid.

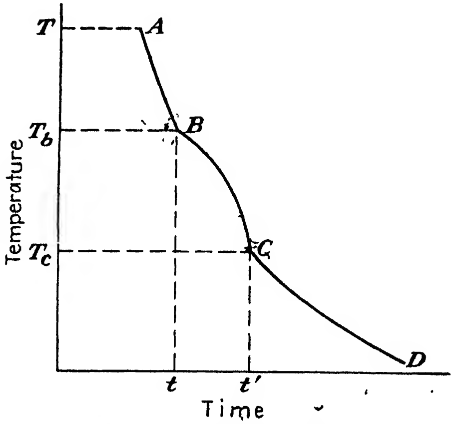


FIG. 3-V.—Typical cooling curve of a binary alloy whose components form a solid solution.

When such an alloy is examined microscopically, it is found that there is no evidence of the existence of either of the separate metals. Alloys of this type, which exhibit cooling curves as in Fig. 3-V, are known to be solid solutions; *i.e.*, the two component metals that are completely soluble in each other in the liquid condition remain completely soluble in the solid state.

In order that the mechanism of the formation of solid solutions may be studied, a given alloy will be assumed that contains a certain amount of a metal *A* of relatively low melting point alloyed with another given amount of a metal *B* of a higher melting point, and it will also be assumed that *A* is the solvent and *B* the solute. When solidification starts, the solid that separates first contains less of *A* than does the liquid alloy as a whole. The remaining liquid metal is thus left richer in *A* than it originally was, and it, therefore, possesses a lower melting point. Because of this varying composition, the solidification of the alloy will cover a wide and falling range of temperatures. Otherwise, if the crystals formed were of the same composition as the liquid alloy, solidification would take place at a constant temperature as in a pure metal, etc. As soon as the temperature falls, more solid is formed containing more *A* than the first solid to solidify, but less of *A* than the liquid metal from which it solidified. This process of solidification will continue until the whole of the alloy becomes solid. During the stages of solidification, therefore, the alloy consists of cores of solid containing least *A*, shells of solid surrounding them containing more *A*, and liquid still richer in *A*. This is an unstable condition; but it is found that, as the temperature is lowered slowly through the solidification range, diffusion takes place within the crystals so that they tend to approach a homogeneous condition. This condition is brought about if sufficient time has been allowed for complete diffusion; otherwise, the unstable condition will persist.

Two Metals Completely Soluble in Both Liquid and Solid States.—In the case of the class of alloys formed by metals that are mutually soluble in the solid and the liquid state, their cooling curves may be grouped together to form the equilibrium, or constitutional, diagram. This has been done in Fig. 4-V by uniting the points *L* and *S* of the independent cooling curves for a series of alloys of metals *A* and *B*. The independent cooling curves have been omitted in Fig. 5-V and the coordinates now become composition and temperature. The figure then becomes a typical curve for this class of alloys. It is composed of two branches, the one *MLM'* is known as the *liquidus*, as any alloy of the series is molten above the line; and the other *MSM'* is known as the *solidus,* as any alloy of the series below the line is entirely solid. Within the area *MLM'SM* the alloys are partly liquid and partly solid. For further description of the solidification of such alloys, an alloy melt of composition *X*, on the diagram, will be considered, which represents 50 per cent of the low-melting metal *A* and 50 per cent of the higher-melting metal *B*, and its cooling from the molten state to room temperature will be followed. An alloy of composition *X* at temperature *T* is entirely liquid, since its condition is represented by a point situated above the liquidus. Upon cooling from *X* to *l*, and the

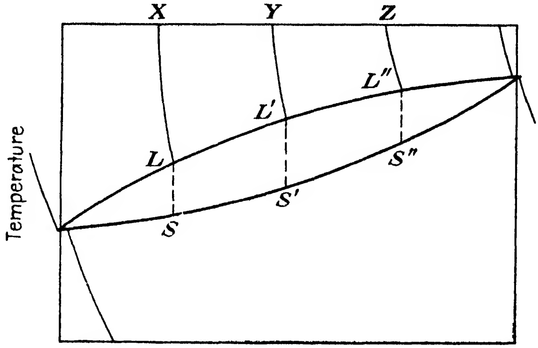


FIG. 4-V.—Diagram showing the construction of the equilibrium diagram for this system.

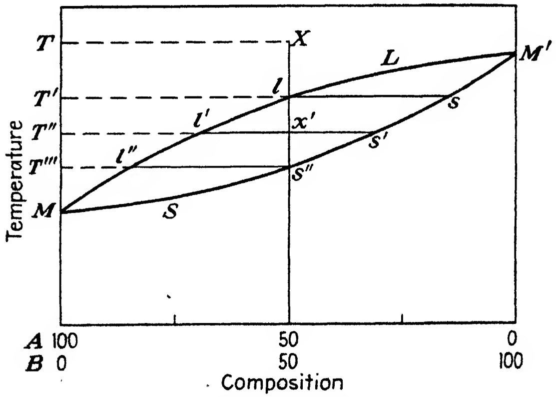


FIG. 5-V.—Typical equilibrium diagram of the binary alloys whose component metals form solid solutions.

temperature dropping from T to T_1, the alloy remains molten. At l, temperature T_1, the first crystals start to solidify. Their composition may be determined by projecting a horizontal line from l to the solidus, intersecting it at s. This intersection s represents the composition of the only crystals that can be in equilibrium at the temperature T_1 with the liquid of composition l.

As the alloy is cooled from l to x', the temperature dropping from T_1 to T_2, the crystals grow by the gradual deposition of the crystallizing alloy, becoming progressively richer in A. The deposited crystalline matter has a varying composition, varying along the solidus from s to s' and the liquid in equilibrium varying in composition along the liquidus from l to l'. At point x', therefore, the composition of the crystals in equilibrium with the liquid melt l', at the temperature T_2 is represented by the point s'. In order to have this condition, the cooling must be sufficiently slow between temperatures T_1 and T_2 to allow proper diffusion in each crystal, so that the concentric layers of varying composition assume the same composition s' and become homogeneous. This condition takes place readily during the crystallization of liquid solutions with sufficient slow cooling, but will not take place in the solid state; *e.g.* if the cooling has been relatively fast, the crystals will remain heterogeneous (parts having different properties), *i.e.*, the layers of crystalline matter will not be of identical composition. At s'', when the temperature T_3 has been reached, crystallization will be complete. The last liquid to solidify will have a composition l''.

If a microscopic examination is made of the structure of a copper-nickel alloy (these two elements are completely soluble) in the unstable condition, it would appear as in Fig. 6-V, *A*. The white dendrites or cores are richer in nickel than the black surrounding material, which contains more copper. This is graphically shown in Fig. 7-V, *A* and *B*, in which the dendrites (white in the figure) are of the composition s and the surrounding material, which is precipitated on further cooling, a composition denoted by point s'. This latter material will not only surround the original dendrites but will also tend to form dendrites of that composition. Thus, the solidification might be considered as taking place in steps. Actually, it is a continuous process, each precipitation grading gradually from the previous one. The reason for the abrupt boundary between the white and the black areas is that the acid attacks and blackens all nickel-copper alloys up to a certain limit of nickel content (corrosion limit) and leaves unattacked the central cores, which are richer in nickel.

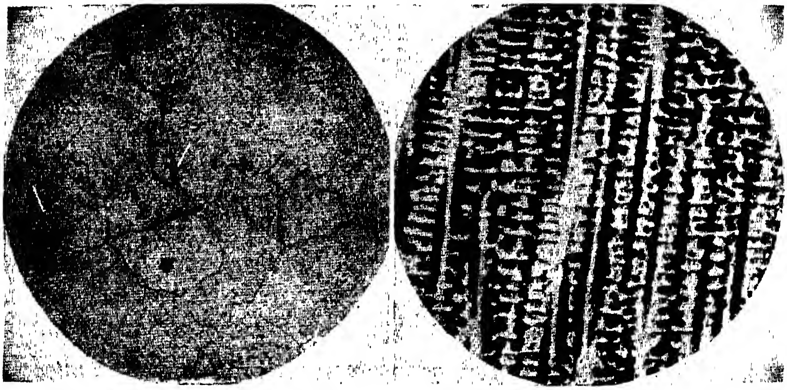
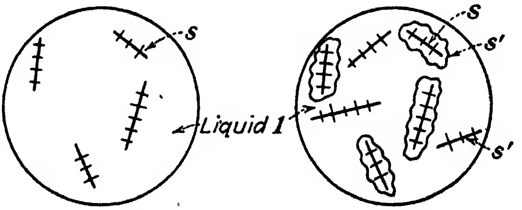

A B

FIG. 6-V.—*A*, nickel-rich core in a copper-nickel alloy before annealing. *B*, homogeneous solid solution after annealing. (*From G. E. Doan, "Principles of Physical Metallography."*)

In order to develop a state of equilibrium in this alloy, it is necessary to utilize one of the following procedures: (1) heat for a long time just below the melting point and then cool as desired; (2) hot-work the alloy; or (3) cold-work it and reheat to just below the melting point. Any one of these procedures will bring about complete diffusion of the cores and surrounding material so that the microstructure will now appear as a homogeneous solid solution (Fig. 6-V-*B*), in which we cannot detect,



A B

FIG. 7-V.—*A*, dendrites of composition *S* in Fig. 4-V. *B*, surrounding material of composition *S*, and dendrites of composition *S* in Fig. 4-V. (*After G. E. Doan.*)

by the microscope, the individual grains of nickel and copper. The grains will all appear polyhedral in shape, like those of a pure metal.

Industrially this homogeneous condition is not accomplished by very slow cooling during solidification, as this would induce an undesirably large grain size and require a long time for the diffusion. In simple steels, which are of this type, the equalization by diffusion is brought about by reheating, soaking, and hot-working the cast metal.

At any temperature T_2 the composition of the solid and the liquid may be found by drawing a horizontal reference line through x' from T_2, so that it cuts the liquidus at l' and the solidus at s' (x' in this case is the reference point). The position of the point l' and the point s', referred to the composition scale, gives the relative proportion of A and B in the liquid and the solid metal. Thus l' corresponds to approximately 70 per cent of A and 30 per cent of B, while s' corresponds to approximately 30 per cent of A and 70 per cent of B.

In addition to the composition of the two phases present at any given temperature on the diagram, there can also be determined the relative amounts of each of the two phases for any given original composition. The quantity of each phase may be found by drawing a vertical line from the point on the composition line that corresponds to the composition in question, as, for example, the vertical line denoting the $50A50B$ composition in Fig. 5-V. The point where the vertical line intersects the temperature horizontal line may then be considered the fulcrum of a lever system. The relative amounts of the phases are then determined by the use of the lever system, or "reaction," which states that the relative amounts of the phases are inversely proportional to the lengths of the lever arms. As an example of the use of the lever reaction, let us determine the amount of solid and liquid present for composition $50A50B$ (Fig. 5-V) at temperature T_2. At this temperature, the length of the line $x's'$ represents the amount of liquid; the line $x'l'$, the amount of solid; and the total line $l's'$ the length of the lever system. In making the calculations, the proportions must be set up with respect to one or the other of the phases. In this example, the conditions are set up with respect to A.

$$\text{Melt } x = 50A \qquad l' = 70A \qquad s' = 30A$$

$$x's' = 50 - 30 = 20$$

$$x'l' = 70 - 50 = 20$$

$$l's' = 70 - 30 = 40$$

$$\text{Amount of liquid at } T_2 = \frac{x's'}{l's'} = \frac{20}{40} \times 100 = 50 \text{ per cent}$$

$$\text{Amount of solid at } T_2 = \frac{x'l'}{l's'} = \frac{20}{40} \times 100 = 50 \text{ per cent}$$

The above procedure, which gives both the composition of each phase and the relative amount of each of the two phases present, can be applied in any binary system, regardless of whether the phases be two liquids, part solid and part liquid, or two crystalline phases in a solid alloy.

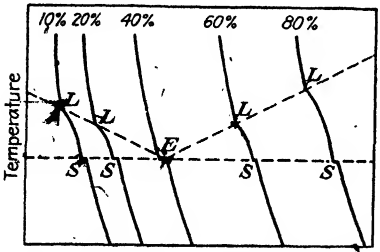


Fig. 8-V.—Diagram showing the construction of the equilibrium diagram for this system.

Two Metals Completely Soluble in the Liquid State and Completely Insoluble in the Solid State.—It is evident that, if these two metals are insoluble when solid, they must crystallize separately into distinct crystals. Thus, a microscopic examination should reveal that the solid alloy is an aggregate of the two metals.

In Fig. 8-V the cooling curves of a number of alloys of the same series of this class of alloys have been grouped together to form the equilibrium diagram. Alloy melts of 10, 20, 40, 60, and 80 per cent of metal B have been selected. It can be seen, therefore, that, with the exception of the alloy containing 40 per cent of B, the solidification, from the cooling curves, begins at L, is retarded from L to S, and completes solidification at E, at constant temperature. These points have been united and the equilibrium diagram has been formed, the coordinates now being composition and temperature, Fig. 9-V. Now consider the features of the solidification of these alloys. The alloys begin to solidify at different temperatures according to their composition. With respect to the per cent of metal B, therefore, with compositions 0 to 40, the solidification temperature is lowered, L to E, while with compositions 40 to 100 the temperature is raised. The final solidification of all alloys takes place at a constant temperature *S*. Obviously, then, *LEL'* is the liquidus line and *LSES'L'* is the solidus. The alloy containing 40 per cent of *B* obviously has the lowest solidification temperature of all in the series. This alloy of lowest melting point is known as the *eutectic alloy* and the point *E* as the *eutectic point.* It is evident from the cooling curve that, like pure metals, eutectic alloys solidify at constant temperature, the eutectic temperature. Eutectic alloys are known to be aggregates of the metals generally finely divided.

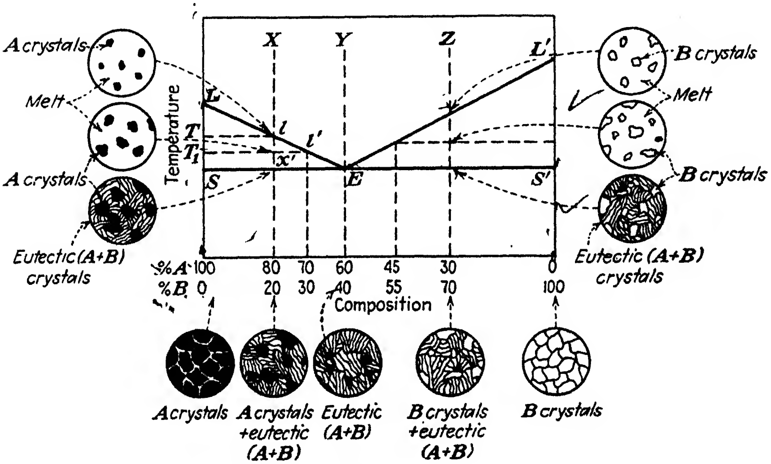


FIG. 9-V.—Typical equilibrium diagram of the binary alloys whose component metals are insoluble in the solid state and the structures of the alloys *X*, *Y*, and *Z*, during solidification and at room temperature.

The first alloy *X* to be considered is a homogeneous liquid and remains in that condition until the temperature has fallen to *T* at the point *l*, the intersection of the vertically drawn line with the liquidus *LE*. Crystallization now starts and, since the two metals are insoluble in each other, and since metal *A* is predominant, the first crystals to separate out will be crystals of pure metal *A*. At any temperature T_1, the composition of the solid and the liquid in equilibrium is found by drawing a horizontal from the solidus through *x'* until it meets the liquidus line. In this case, it intersects *LS* at T_1, which is pure metal *A*, and intersects *LE* at *l'*. Referring to the scale of composition, the relative amounts of *A* and *B* in the liquid will be 70 per cent of *A* and 30 per cent of *B*. The solid, of course, remains pure *A*. The relative amounts of solid and liquid in this case being inversely proportional to the segments through x' are, $x'T_1$, which equals $66\frac{2}{3}$ per cent liquid, and $x'l'$, which equals $33\frac{1}{3}$ per cent of solid. As the temperature is lowered, crystals of pure *A* are crystallizing from the remaining liquid, which becomes progressively richer in metal *B*. The proportion of *B* in the liquid increases until the limit of saturation is reached for *B*, which then begins to crystallize. Since the liquid is already saturated with respect to *A*, any crystallization of metal *B* will necessitate a crystallization of metal *A* in order that the liquid composition may remain constant and not become supersaturated with respect to either metal. Therefore, the two metals will crystallize simultaneously at a constant temperature until there is complete solidification, the composition of the liquid remaining constant while it exists. *E* is, therefore, the eutectic point.

The relative amounts of *A* crystals and of eutectic in alloy *X* may be determined by using the lever system. The calculations are as follows:

$$\text{Amount of eutectic} = \tfrac{20}{40} \times 100 = 50 \text{ per cent}$$
$$\text{Amount of } A \text{ crystals} = \tfrac{20}{40} \times 100 = 50 \text{ per cent}$$

For an alloy of composition *Y*, the liquid metal will remain homogeneous until a temperature is reached that corresponds to *SES'*, where solidification will take place at constant temperature. The relative amounts of *A* and *B* in the solid eutectic alloy can be determined by the use of the lever as before; however, the calculations, in this case can be made by observation. The amount of *A* = 60 per cent and the amount of *B* = 40 per cent.

For an alloy of composition *Z* the process is identical with that followed for *X*, except that the solid that first separates is pure metal *B*. On cooling, *B* separates out, while the composition of the liquid follows *L'E* and becomes increasingly rich in *A*. When the composition and temperature of the liquid have reached *E*, there will be simultaneous crystallization of *A* and *B* as the eutectic.

Since all alloys in the area *LES* contain less of metal *B* than the eutectic, they can conveniently be referred to in future diagrams as *hypoeutectic* alloys; all alloys in the area *L'ES'*, containing more of metal *B* than eutectic, can be referred to as *hypereutectic* alloys; and an alloy of eutectic composition, as a *eutectic* alloy.

From the discussion and from the typical structures drawn at the bottom of Fig. 9-V, three types of structures can be expected from alloys *X*, *Y*, and *Z*: (1) the structure of hypoeutectic alloys, as being composed of primary crystals of *A* surrounded by eutectic; (2) the structure of eutectic alloys, consisting of a finely divided aggregate of minute particles of both metals; and (3) the structure of hypereutectic alloys, composed of primary crystals of *B* surrounded by eutectic. Eutectic alloys frequently occur structurally as thin alternate and parallel plates of each of the two constituents as shown and, in some cases, consist of rounded or elongated particles of one of the constituents embedded in the matrix of the other.

Two Metals Completely Soluble in the Liquid State and Partially Soluble in the Solid State.—It is seldom that metals are absolutely insoluble in each other in the solid state, each metal generally being capable of retaining a small percentage of the other metal in solid solution. Certain modifications must, therefore, be made in the equilibrium diagram for this class of alloys. Fig. 10-V shows a typical equilibrium diagram with the cooling curves omitted. *LEL'* is the liquidus, *LSES'L'* is the solidus, while *SM* and *S'M'* are branches of the curve of solubility of the solid metal, indicating the changes of solubility of the metals as the alloys cool from the eutectic to atmospheric temperature. It can be seen that the *S* of the eutectic line now corresponds with the saturated solid solution of *B* in *A*, and *S'* with that of the saturated solid solution of *A* in *B* at eutectic temperature.

Further, by projecting to the composition line, it is shown that *A* retains approximately 15 per cent of *B* at the eutectic temperature and decreases in amount with decreasing temperature until, at atmospheric temperature, *A* retains 10 per cent of *B*. Similarly, at the eutectic temperature, *B* retains 10 per cent of *A* and decreases to 5 per cent at atmospheric temperature.

The process of solidification of the three alloys *X*, *Y*, and *Z* will now be considered.

The alloy *X* starts to crystallize at temperature *T*, with the first crystals having the composition as determined by dropping a vertical to the composition line from the intersection of the horizontal through l with the solidus. As the temperature drops, other crystals form. The new solid differs from the old in that the precipitated solid is becoming richer in B and the liquid in equilibrium with the solid is also becoming richer in B. This, of course, postulates slow cooling—slow enough for proper diffusion. Solidification thus continues until temperature T_1 is reached; whereupon there is complete solidification. The solidified metal at room temperature will consist simply of a solid solution of B in A. In this particular melt, A is undersaturated with respect to B and remains undersaturated over the entire temperature range, from T_1 to atmospheric.

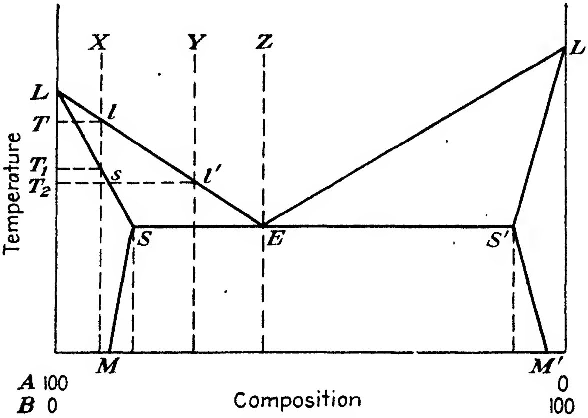


FIG. 10-V.—Typical equilibrium diagram of the binary alloys whose component metals are partially soluble in the solid state.

The alloy Y starts to solidify at the temperature T_2, where the first crystals to solidify have a composition s. On gradual cooling, this solid reaches the composition S, which point represents the maximum solubility of B in A. The liquid in the meanwhile has changed in composition from l' to E, at which temperature it solidifies at constant temperature, as the eutectic alloy. The eutectic is formed by simultaneous deposition of the crystals of the saturated solid solution of B in A and A in B. Just following complete solidification, the alloy is considered to be an aggregate of the two solid solutions. The diagram indicates, however, that the mutual solubilities of the metals A and B decrease, which would mean that each crystal must undergo some transformation of structure. The transformations are indicated by the lines *SM* and *S'M'*. At atmospheric temperature, therefore, the alloy has become an aggregate of two solid solutions whose compositions are represented by points *M* and *M'*. The alloy would be composed of free crystals of solid solution *M* and eutectic, whose components are solid solution *M* and solid solution *M'*. Thus, alloys between the compositions of *M* and eutectic would be composed of free saturated solid solution of *B* in *A* and eutectic, while those alloys between compositions eutectic and *M'* would be free saturated solid solution of *A* in *B* and eutectic.

The alloy *Z* cools down to point *E* without solidification of any solid. At *E* the whole mass solidifies at constant temperature, which is entirely eutectic, composed of an aggregate of the saturated solid solutions *B* in *A* and *A* in *B*. The compositions of the two solid solutions change as the temperature falls in a manner as is indicated by the solubility curves *SM* and *S'M'*, and the aggregate has a change in structure.

Binary Alloys That Show the Formation of Definite Chemical Compounds.—When two metals such as *A* and *B* are capable of forming one or several definite chemical combinations, there is obtained a diagram much more complex than those previously studied. For simplicity, consideration will be given to the formation of only one such combination, although many instances are known in which more than one compound is formed.

The simplest method of conceiving an intermetallic compound from the viewpoint of cooling curves, constitutional diagrams, and microstructure, is to regard it as a pure metal; as it is, in fact, a pure substance of metallic character. A system of alloys of two metals in which such compounds occur is simply a series of other simple systems. From the equilibrium diagram, Fig. 11-V, which represents the class of alloys that are completely soluble in the liquid state and completely insoluble in the solid state and form a compound, it is evident that it may be broken up into two systems, both of which have just been discussed. Assuming that the metals *A* and *B* do form a compound, the first system would be between *A* and *C* (A_xB_y), and the second, between *B* and *C*. Combining the two diagrams, we see that *C*, which is the solidus of the individual diagrams, represents the meeting point of the liquidus and the solidus of both diagrams and represents now the freezing point of the compound. In a complete equilibrium diagram, the presence of a pure compound will be marked by (1) a maximum peak in the liquidus, (2) a meeting point of the solidus and the liquidus, and (3) a chemical analysis that is found to contain the metals in approximately simple proportions. The microstructure of the intermetallic compound may be represented by the typical structure of a pure metal, *viz.*, an aggregate of crystals all of one kind. It becomes quite difficult at times, however, to distinguish between these crystals and those of a slowly formed solid solution.

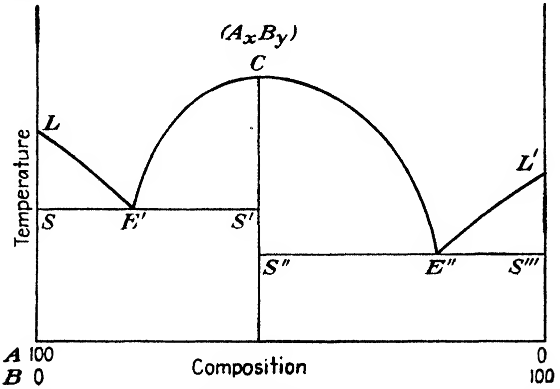


Fig. 11-V.—Typical equilibrium diagram of a binary alloy whose component metals are insoluble in the solid state and form a compound.

Two Metals Completely Soluble in the Liquid State and Partially Soluble in the Solid State, with the Occurrence of a Peritectic Reaction.—Figure 12-V shows the equilibrium diagram for this particular class of alloys. *LPL'* is the liquidus and *LHKL'* is the solidus, with the lines *HM* and *KN* representing branches of the solubility curve. The two branches of the liquidus curve intersect at point *P*, which is not a eutectic point. At the temperature *P* there are two solids *H* and *K* in equilibrium with the liquid of composition *P*, with the following reversible reaction taking place, depending upon whether the system is being heated or cooled.

$$\text{Solid } K \rightleftarrows \text{solid } H + \text{liquid } P$$

For convenience in the discussion, the different phases in the diagram have been lettered, *viz.*, α the first solid solution to separate, and β, the second solid solution.

Upon cooling any melt between *H* and *K*, the first crystals to separate would be primary α solid solution, which would vary in composition along the solidus *LH* to *H*, the liquid varying along the liquidus *LP* to *P*. At the temperature *HKP* there is α solution *H*, of *B* in *A* and liquid *P* with a proportion of *B* greater than that of *H*. The new solid solution, β*K*, of *A* in *B*, which is also in equilibrium with the liquid, starts to crystallize. This crystallization would make the liquid slightly richer in *B*, but this temporary excess diffuses into the previously formed α crystals in sufficient quantity to maintain equilibrium and allow the liquid eventually to solidify as crystals of composition *K* or β solid solution. The completely solid metal then consists of crystals of solid solution α intermingled with crystals of solid solution β, some of which were formed by the reaction between the primary α crystals and the liquid, while the remainder were deposited as such from the liquid.

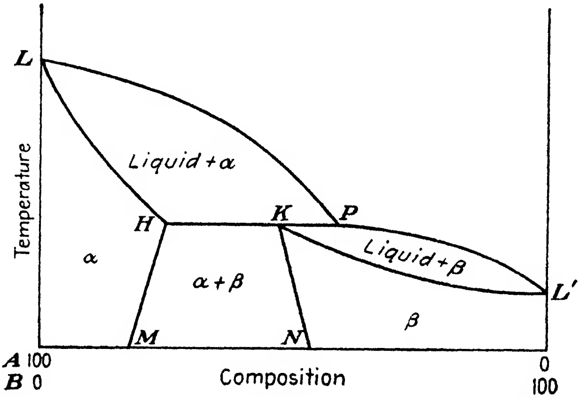


FIG. 12-V.—Typical equilibrium diagram of a binary alloy whose component metals are partially soluble in the solid state with the occurrence of a peritectic reaction.

Upon cooling any melt between *K* and *P*, the first crystals to separate would be primary α, which changes in composition to *H*, the liquid varying to *P*. At the temperature of *HKP* the solid of composition *H* entirely disappears and is replaced by β of composition *K*. Solidification is not completed at this point, for the proportion of *B* in the liquid is greater than that which corresponds with composition *K*. As the temperature falls, the liquid, therefore, varies along *PL′*, while the solid in equilibrium varies along *KL'*. Solidification continues as in the case of a single solid solution, in this case β and liquid, the final solid being β.

The Recrystallization of Solid Solutions.—In Fig. 5-V it has been shown that the conditions of solubility are quite the same whether the substances are liquid or solid. Such a similarity carries much further, in that any of the changes taking place during the solidification of a liquid alloy may take place when the alloy is completely solid; *i.e.*, there may be, in alloys that are completely solid, a secondary recrystallization that is an analogy of primary crystallization.

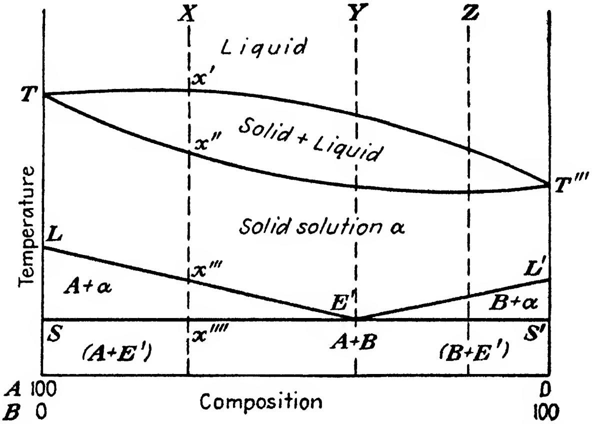


FIG. 13-V.—Typical equilibrium diagram of binary alloy whose component metals are completely soluble at high temperatures and completely insoluble at lower temperatures.

Such a case is shown in Fig. 13-V, in which the primary solid solution is resolved into secondary recrystallization with a eutectoid. The upper part of the diagram is the same as the alloy series of Fig. 5-V. At a temperature below complete solidification, a secondary crystallization takes place with the precipitation of either pure *A* or pure *B*, depending upon the alloy composition. Crystallization proceeds as in the series of alloys of Fig. 9-V, with a lowering of temperature and precipitation of the pure phase; the solid approaches and reaches the point *E'*, which has a significance similar to that of *E* (Fig. 9-V). Thus, at *E'* the two separating solids change simultaneously to a complex mixture of the two new constituents, in the case *A* and *B*, to which the name "eutectoid" has been applied.

For a more thorough study of the process, the crystallization of melts *X*, *Y*, and *Z* will be followed in detail. The liquid melt *X* begins to solidify at a temperature corresponding to the point x', and becomes completely solid at x'' as a solid solution that we shall call *alpha* α. When the temperature has dropped to x''', pure metal *A* starts to precipitate from the solid solution α and continues as the temperature drops. Upon precipitation of *A*, the remaining solid solution changes, with decreasing temperature, in composition from x''' to E'. At the temperature x'''', the remaining solid solution, which now has the composition E', precipitates as a complex mixture of metals *A* and *B*, the eutectoid.

The liquid melt *Y* begins to solidify upon reaching the liquidus and becomes a uniform solid solution at the solidus and makes no further change until the temperature E' is reached. The solid solution then changes, at constant temperature, to the eutectoid mixture of *A* and *B*.

The liquid melt *Z* follows steps of transition similar to those of melt *X*, with the exception that the first component to separate from the α solid solution will be *B* instead of *A*.

The structures of the melts will be very similar to those discussed under Fig. 9-V. Therefore, the hypoeutectoid structure will be composed of crystals of *A* surrounded by eutectoid; the eutectoid structure, of a finely divided aggregate of both metals; and the hypereutectoid structure, of crystals of *B* surrounded by eutectoid.

In this discussion of Fig. 13-V, it has been assumed that the transition of the solid solution α is a pure metal, but this is not always the case, as, frequently, the new constituent may be a compound or a solid solution whose compositions vary with temperature. Since the transitions are taking place in the solid state, ordinary rates of cooling do not usually allow the separating component to crystallize in shapes as the primary crystals from a liquid solution and, hence, often aid in distinguishing secondary crystallization. Slow cooling frequently permits the secondary crystallization to assume primary forms.

EXPERIMENTAL PROCEDURE

The methods used in determining critical points of metals and alloys are (1) plotting heating and cooling curves on time-temperature coordinates; (2) dilatometric measurement, which involves plotting the progressive expansion or contraction on linear-expansion temperature coordinates; (3) plotting the change in ferromagnetic characteristics on a temperature magnetic-intensity diagram; or, (4) study of the changes in microstructure.

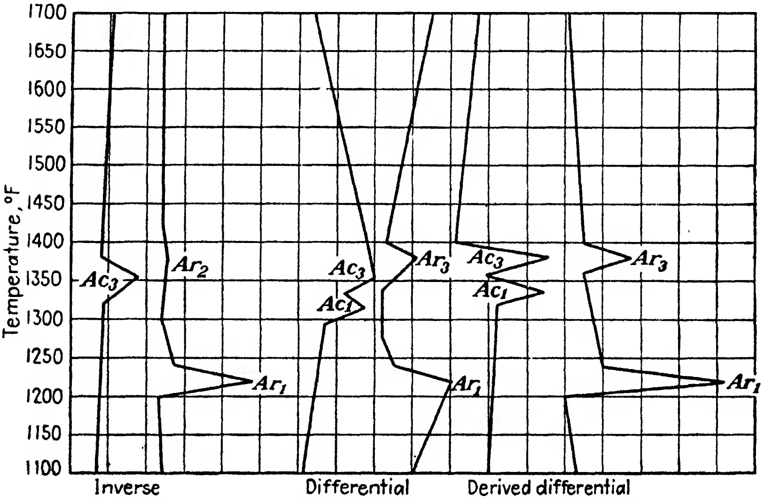


FIG. 14-V.—Heating and cooling curves for a 0.50 per cent carbon steel.

Heating and Cooling Method.—A complete time-temperature heating and cooling curve is produced by heating the metal from room temperature to above the melting point or cooling it from above the melting point to room temperature and noting the temperature of the metal at regular, predetermined intervals of time. The temperature readings are taken at equal time intervals of from 1 to 3 min., depending upon the rate of heating or cooling.

The method consists in heating the metal charge in a crucible placed in a furnace capable of heating the charge uniformly to the highest temperature required and so insulated that the charge can be cooled at a moderate, uniform rate. The temperature readings are taken with a thermocouple with a working range within the limits of the temperature range to be studied. The time intervals may be made by the use of a stop watch.

It is found that, in plotting the points determined by the above method, the thermal points that indicate heat evolution or heat absorption are not very conspicuous at times. In fact, the slope of the curve may be so gradual that it will be impossible to select the correct transformation temperature. The three methods used to overcome this difficulty are illustrated in Fig. 14-V, which shows respectively, the inverse, the differential, and the derived-differential heating and cooling curves for a 0.50 carbon steel. A comparison between the intensity of the thermal arrests can easily be made.

Inverse-rate Method.—The procedure followed in this method is essentially the same as in the simple time-temperature method. The time, however, must be more accurately measured.

The curve is obtained by plotting the absolute time interval required for a definite and predetermined temperature change. The lower temperature of each successive temperature interval is plotted against the time.

Example: If on cooling it requires 50 sec. for the temperature to drop through a predetermined temperature interval of 5°F. from 1225 to 1220°F., the 50-sec. interval is plotted against 1220°F., etc.

Differential Method.—The greatest irregularity arising in the simple determinations is in the quality of the furnace equipment. The importance of the relation between the control of the furnace and accuracy of determining the transformation points is evident when we consider that the heat effect in the metal is opposite in effect to the furnace, at the critical. Heat is being evolved by the metal's tending to raise its temperature while the furnace is cooling and is attempting to lower the metal's temperature. The result of these tendencies is to mask the transformations. To eliminate this factor, a neutral or non-transforming metal, such as platinum or austenitic nickel steel, is heated or cooled along with the sample. If the sizes and specific heats of the two specimens are approximately the same, the temperature difference between the two will be small except at the transformation point of the specimen and the effects of the rate of heating and cooling will be negligible.

At a transformation point—using an experimental setup, as is shown in Fig. 15-V, in which there are two thermocouples, one a differential couple with hot junctions in both specimen and a neutral body and the other in the specimen to indicate its temperature—there will be a temporary arrest of the temperature drop in the specimen, while the temperature of the neutral body will continue to fall without interruption. Minor transformation points will thus appear as sharp peaks and will be easily detected.

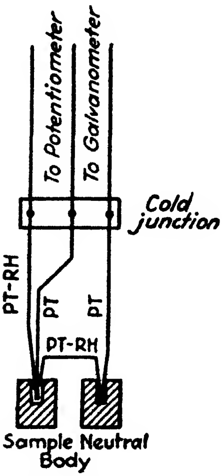


Fig. 15-V.—Experimental arrangement of the thermocouples, sample, neutral body, for obtaining differential or derived differential curves. (*From Nat. Bur. Standards Tech. Paper*, 170, 1921.)

This so-called *differential* method consists in plotting the temperature of the specimen against the temperature difference between the specimen and the neutral body.

Derived Differential Method.—This method of investigating the heating and cooling characteristics of metals is another means of intensifying transformation points that otherwise would not be detected. The experimental procedure is the same as used in the differential method except the manner in which the data are recorded.

In the recording of data and the construction of the curve, the method consists in plotting successive drops in temperature against the corresponding rate of cooling for each degree of temperature. Thus, if in cooling from 1450°F., the temperature difference between the specimen and neutral body for every 5°F. drop in specimen temperature was found to be 4, 6, 9, and 14°F., these values would be plotted against 1450, 1445, 1440 and 1435°F., respectively. In plotting heating curves, the difference values are plotted against the lower temperature value of the interval.

Dilatometric Method.—This method of investigating the transformation characteristics of metals and alloys is based on the fact that, in most metals and alloys, the transformations occurring in the solid state are accompanied by expansions or contractions. By the use of a sensitive measuring device, the linear dilation of the metal at its transformation points can be detected and correlated with temperature. Dilatometric analysis, it should be noted, can be used only for the determination of transformations that occur in the solid state and well below the melting point of the material being investigated.

The principle of the dilatometer is shown in Fig. 16-V. It is essentially a mechanism that measures and magnifies the expansions and contractions of a sample so that they can be easily studied for any transformation changes that may occur. The specimen *S* is rigidly supported on an alloy base having a nearly uniform and linear expansion. A quartz rod *R* contacts the top of the specimen and extends through the top of the furnace *F*, where it contacts and actuates a lever *L*. As the lever is moved, it actuates a stylus *P* in a horizontal direction on a chart *C*, which rotates at a constant speed, and traces a time-expansion curve of the specimen. With the use of such auxiliary equipment as a thermocouple and a pyrometric recorder, the ordinate of the curve can be made to read in temperature rather than in time. This gives typical time-temperature heating and cooling curves, from which can be obtained the temperature of the transformation points of the specimen.

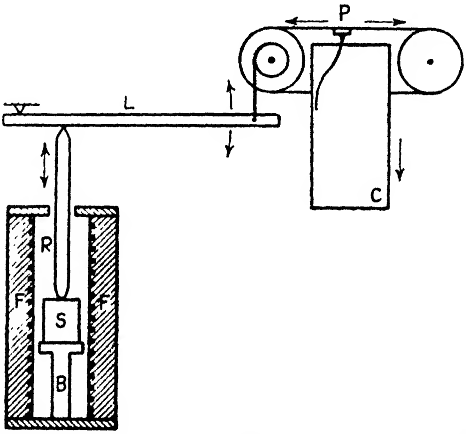


FIG. 16-V.—Principle of the simple mechanical dilatometer. (*From G. L. Kehl, "The Principles of Metallographic Laboratory Practice,"* 1939.)

Magnetic Method.—The magnetic method for the determination of transformation points depends upon the loss of magnetic properties at temperatures where the transformations occur.

These points may be determined in a rough manner by inserting the specimen in a suitable furnace and noting the temperature at which it loses its attraction for a magnet. More accurate methods may be employed, such as placing a round rod of the material securely between two similar-sized rods of ingot iron or wrought iron, inserting a thermocouple within the specimen and placing the composite within a suitable furnace. The ingot-iron rods are of such a length that they extend beyond the ends of the furnace. A magnetic flux is induced in the specimen by a coil or an electromagnet placed around one end of the composite rod, while a galvanometer arrangement is connected to the other end. When, on heating, the temperature is reached at which the magnetic flux ceases to pass through the sample, as will be indicated by the galvanometer, the loss of magnetism indicates the critical point. In the course of cooling, the critical point will be indicated by the sample's regaining its magnetic properties.

Microstructure.—Transformations may also be investigated in metals and alloys by a study of the microstructure of a series of specimens that have been quenched from temperatures progressively nearer the transformation temperature. By drastic quenching, the structure at the quenching temperature can be retained and its progress studied from the microstructure. This method is particularly adaptable to the study of solid-solubility lines or to conditions where it is difficult to establish true equilibrium.

Suggested Questions and Problems for Study and Class Discussion

Copper melts at 1083°C., silver at 960.5°C. A eutectic is formed at 28.0 per cent copper, the temperature being 779°C. The solid solubility of copper in silver at the eutectic temperature is 9.0 per cent, while silver in copper is 8.0 per cent. The solubility of each is approximately 1.00 per cent at room temperature.

1. Draw the constitutional diagram, labeling all points and fields.
2. In an alloy of 60 per cent Cu
   - *a.* What is the approximate temperature of initial solidification?
   - *b.* When the temperature is 20°C. below the point of initial solidification, what is the composition of the liquid phase and what proportion of liquid phase is present?
   - *c.* When the temperature is 10°C. above the point of final solidification, what is the composition of the liquid phase and what proportion is liquid and what solid?
   - *d.* Describe the changes during cooling from the point of initial solidification to room temperature.

3. Sketch the general appearance of the microstructure and label each constituent of the following four melts of a copper-silver alloy: 5.0, 15.0, 28.0, and 70 per cent copper.

4. Calculate the amount of eutectic in the above four alloys at room temperatures.

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CHAPTER VI

THE IRON-CARBON DIAGRAM

The study, that has been made in the previous section, on the general theory of alloys will now be used as a basis for the development of the theories regarding iron and steel. The study of the alloys of iron in general is a difficult task, because the equilibrium conditions of the various kinds of steel and cast iron are only incompletely known. Although much fine development work is being done, the investigators are hampered by the many unusual characteristics and by the behavior exhibited by the metal. All this in one way or another contributes to the difficulties in the way of arriving at a true picture of equilibrium conditions. Some of the factors that must be considered in such studies are these: That iron exists in at least three allotropic forms, each form having a definite range of stability and each limiting range varying with the percentage of foreign element present; that certain transformations in the solid state are complete only after long periods of heating, followed by slow cooling; that the type of gas used during these prolonged heatings, whether it be oxygen, hydrogen, nitrogen, or some other, has a definite effect upon the properties of the metal; that high temperatures are needed in most of the work; and that it is difficult to obtain iron alloys free from certain elements and common impurities. In fact, the reader should bear in mind that all commercial alloys are very complex mixtures containing possibly a half dozen elements, all of which have an appreciable effect.

IRON

Iron is the principal constituent in all steels, cast irons, wrought irons, and other ferrous alloys, which play so important a part in the progress of our modern civilization. Iron forms a large number of alloys, both simple and complex, which are of such importance that the materials with which metallurgy deals are divided into two classes—*ferrous*, including all those alloys of which iron is the basic constituent, and *nonferrous*, which comprise the remainder.

In the following discussion it must be noted that reference to pure iron indicates a commercial iron, such as electrolytic or ingot iron, which shows only traces of impurities and, from a laboratry viewpoint, of a product containing 99.95 per cent iron, plus a little carbon, oxygen, hydrogen, nitrogen, silicon, sulfur, phosphorus, and perhaps others, which together amount to 0.05 per cent.

Our present knowledge of pure iron has been acquired through a study of thermal, magnetic, microscopic, and X-ray data. These studies have shown that the internal structure of a metal such as iron is definitely related to the physical properties and that the physical properties are dependent upon the structure.

The structure of iron, as of all metals, may be conveniently thought of in terms of the smallest unit particle known, the atom. This unit is so small that it is difficult to get a conception of it. Still, with the assistance of present-day precision apparatus, which places at our disposal facts that make it possible, for our present purposes, to think of the atom of metal as a very minute "ball," every atom being just like every other one. Each atom, or "ball," has a definite weight and exerts strong and definitely directed forces of attraction or repulsion upon all the other atoms in its vicinity. These forces are electric or magnetic in character, the atom itself being composed of these forces under some bondage that holds them together in certain ways, but still lets certain strong attractions or repulsions be exerted on outside atoms. In most of the metals, these forces are so arranged in the atoms that when the metal solidifies the atoms arrange themselves in a definite crystalline structure known as *cubic*.

The result of both microscopic examination and X-ray diffraction analysis of pure iron has shown that its structure is characterized by this cubic arrangement.

If we were to make a thermal study of pure iron by one of the methods previously discussed, there would be indications of transformation points at two different temperatures, 1670°F. (910°C.) and 2535°F. (1390°C.). This must indicate that three different forms of iron exist within the successive temperature ranges. X-ray analysis has shown that from 0°F. to 1670°F. pure iron has a body-centered crystalline structure in which the atoms of the unit cube of iron are located on each corner of the cube with an additional one in the center. This structural arrangement is shown in Fig. 1-VI*A*. From 1670 to 2535°F. the structure is entirely modified and has a face-centered crystalline structure in which the atoms of the unit cube of iron are located on each corner of the cube and in the center of each of the cube faces. This structural arrangement is shown in Fig. 1-VI*B*. From 2535°F. to the melting temperature (2795°F. or 1535°C.), there is another modification.

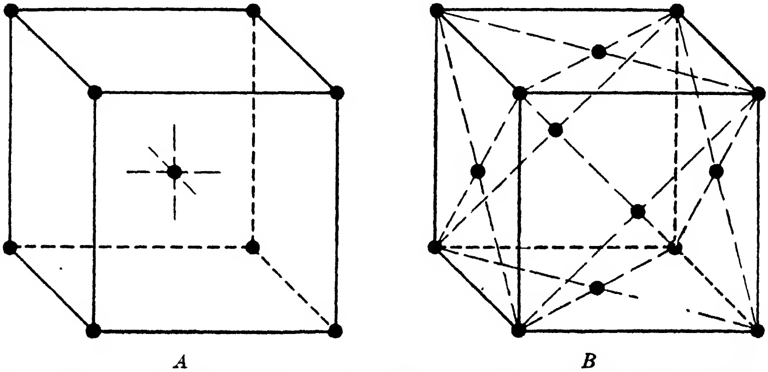

Fig. 1-VI.—*A*, Body-centered cubic lattice. *B*, Face-centered cubic lattice.

The crystalline structure reverts to the body-centered cubic form. These so-called *allotropic* modifications of iron have been termed *alpha, gamma,* and *delta,* according to the order of their appearance on heating. The modification, its crystal structure, and its range of stability are given in Table 1-VI.

Table 1-VI

Modification	Crystal structure	Range of stability	
		°F.	°C.
Alpha	Body-centered cubic	1670 and below	910 and below
Gamma	Face-centered cubic	1670–2535	910–1390
Delta	Body-centered cubic	2535–2795	1390–1535

As each of these crystalline modifications makes its appearance on heating or cooling through the critical temperatures, there will be complete recrystallization to the form that is stable within that range.

There is one other discontinuous change in the physical properties at or within a slight range of 1420°F. (770°C.). It is at this point that alpha iron loses or gains its magnetic property, depending upon whether it is being heated or cooled. In other words, alpha iron is magnetic, while gamma and delta are nonmagnetic. This discontinuity is not accompanied by a change in crystalline structure or in any corresponding phase transformation.

A microscopic study of the cross section of a sample of pure iron will disclose that it is composed, as are other pure metals, of solid, polyhedral-shaped crystals, termed *crystalline grains*, or just *grains*, in which the network of lines represents the boundaries of the crystal. Since it is known that the structure is cubic in character, we can conceive that each of these grains is made up of cubes of iron in rows, all perfectly aligned with respect to each other. The direction of the lines or planes of formation of the grain is known as its *orientation*. This orientation varies from grain to grain, so that a section through a sample will cut each grain a little differently from its neighbor and may produce the grain contrast on subsequent etching, as illustrated in Fig. 1-II.

The above observation indicates that each individual grain is two-dimensioned, but actually the grains are three-dimensional in character. Thus, each of them is actually made up of rows and layers of cubes of iron, all oriented with respect to each other in the individual grain, but not with respect to the rows and layers in neighboring grains. For a clear concept of the subsequent discussions it must be emphasized that even though the data discussed pertain to plane surfaces, the reactions are actually occurring simultaneously in all three dimensions.

CONSTITUTION DIAGRAM

The constitution diagram of the iron-carbon system (Fig. 2-VI) has been constructed by collecting and plotting thermal data as was done in the formation of the simple binaries previously studied. This diagram, as do all diagrams previously studied and those to be studied in the succeeding pages, deals only with the constitution of the system; *i.e.*, what phases are present and how much of each is present, and not with their structure. In the diagram, temperature is plotted vertically and composition horizontally. Any point on the diagram, therefore, represents a definite composition and temperature, each value being found by projecting to the proper reference axis.

By applying the principles that were studied from the preliminary diagrams, this system becomes comparatively simple.

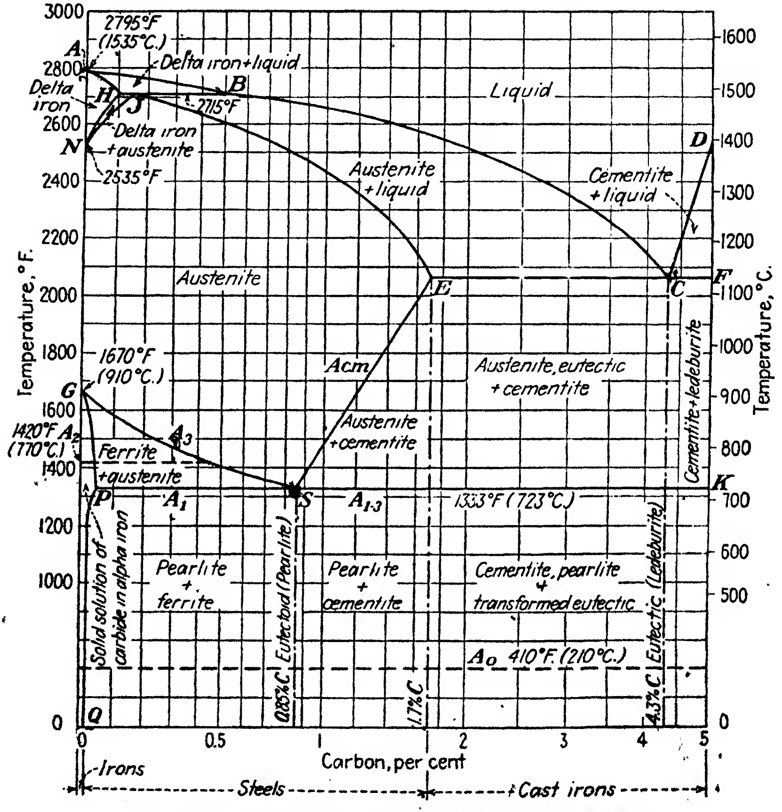

FIG. 2-VI.—Iron, iron carbide constitutional diagram.

The line *ABCD* is the liquidus line, above which all alloys are molten and below which the alloys are partially or completely solid, while *AHJECF* is called the *solidus* as below this line all alloys are completely solid. The area *ABJN* shows the occurrence of a peritectic reaction. If the line *JE* should be extended to *A*, eliminating in that case the area *ABJN*, there could then be set apart *AESKFDC*, which would correspond to a eutectic diagram with partial solubility in the solid state being denoted by *ES*. By drawing a horizontal line through *E*, with line *JE* still extended to *A*, a simple solid solution diagram is formed with recrystallization in the solid state and the occurrence of a eutectoid point.

Knowing that the lines on the diagram represent phase changes and that within the areas bounded by these lines certain phases are present, the next step is to define the phases that are stable within these areas.

Ferrite.—Alpha iron, which exists from a temperature of 1670°F. and below, is found to be capable of holding in solid solution such elements as carbon, nickel, silicon, and phosphorus. It has a maximum solubility for carbon of 0.035 per cent at a temperature of 1335°F. (725°C.), which solubility decreases according to the solid-solubility line *PQ* until at room temperature the solubility is less than 0.01 per cent.

Ferrite is defined as alpha iron (or delta iron) containing any other element in solid solution.

Cementite.—The constitution diagram for the iron-carbon alloys has not been drawn to include the usual range of compositions, since the alloys beyond about 5 per cent have no commercial importance. The actual limitations, however, should have been from 0 to 6.67 per cent carbon. The reason for this limitation is that iron and carbon combine to form an intermetallic compound containing 6.67 per cent carbon, whose formula is Fe_3C. It is found that in the majority of the commercial iron-carbon alloys, which include the steels and white cast irons, the reaction is as if the components are iron and cementite rather than iron and carbon. It is found convenient, therefore, to use this value of 6.67 per cent carbon as the upper limit, as it permits the interpretation of the system from 0 to 100 per cent Fe_3C or 100 to 0 per cent Fe.

Although cementite is one of the fundamental constituents of both steels and cast irons, it is impossible to produce the pure compound in actual practice, because of the modifying effect of the elements either added purposely or contained in the usual commercial compositions. At high temperatures, cementite decomposes into iron and graphite and, therefore, has no observable melting point.

Austenite.—In the previous discussion of the allotropic forms of iron, it was noted that a structure termed *gamma* iron was stable within the temperature range of 1670° to 2535°F., which, as can be seen from the constitution diagram, comprises, on the 0 per cent carbon line, one boundary of the area in which a solid solution exists, known as *austenite.*

Austenite, like other constituents on the diagram, has certain definite solubility relationships, which are dependent upon the temperature and the composition in the area in which it is stable (*NJESG*). Austenite has much the same properties as a liquid solution, in that its original components are merged into a single entity, giving a complete indefiniteness of composition and having entirely new characteristics. Since austenite is one constituent of the iron-carbon constitution diagram and has as one boundary gamma iron, it can be defined as a solid solution of carbon or iron carbide in gamma iron. There is still considerable question whether austenite holds the carbon in solid solution as the element or whether it holds it as cementite. Since a chemical analysis of steel for carbon will be reported as such and not as cementite, reference will always be made to the carbon content of the austenite rather than to its cementite content, unless otherwise noted.

As far as its solubility characteristics are concerned, austenite is capable of holding 1.70 per cent carbon in solid solution at a temperature of 2065°F. Austenite, however, can hold in solid solution anything from 0 to 1.70 per cent carbon, depending upon the temperature. The limits of temperature and composition are shown by the area *NJESG* of Fig. 2-VI.

Pearlite.—With moderately slow cooling, decomposition of austenite containing 0.85 per cent carbon[1] results in the formation, at a temperature of 1335°F., of a constituent known as *pearlite,* which is composed of wavy or curved alternate plates,

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[1] The percentage of carbon in commercial eutectoid steels is considered to be between 0.80 to 0.85 carbon owing to the fact that a microscopic examination shows them to be appreciably free from both free ferrite and free cementite, in other words 100 per cent pearlite.

Ferrite that is not included in pearlite is often called *free, excess, massive, proeutectoid* or *noneutectoid.*

Cementite that is not included in pearlite is often spoken of in the same terms as ferrite.

In the discussion, when necessary, both ferrite and cementite other than that contained in the pearlite will be spoken of generally as *free,* and in a few cases as *proeutectoid.*

or lamellae, of ferrite and cementite. It has been named *pearlite* because it frequently exhibits a display of colors similar to the color display obtained from mother-of-pearl or from the lines of a diffraction grating.

Pearlite is defined as a lamellar aggregate of ferrite and cementite that is chemically and structurally of eutectoid composition. The statement regarding composition is made, inasmuch as the proportion of free ferrite to cementite is approximately 7 to 1. This proportion may be considered the normal structural proportion that is produced under slow furnace cooling.

Ferrite Solubility Curve.—In pure iron the transformation from gamma to alpha iron occurs at a temperature of 1670°F. (910°C.). If varying amounts of carbon, from 0.00 to 0.85 per cent, are added to the gamma iron, the temperature at which the alpha iron or ferrite forms from the gamma iron decreases from 1670°F. to 1335°F. This is indicated on the iron-carbon diagram by the line *GS*, which shows the solubility of ferrite in austenite and is known as the ferrite solubility curve.

Cementite Solubility Curve.—At a temperature of 2065°F. (1130°C.), which is known as the *eutectic temperature*, austenite will hold 1.7 per cent carbon in solid solution. If austenite of this composition is cooled slowly, cementite will precipitate from the solid solution. With the precipitation of cementite (6.67 per cent C) the carbon content of the remaining austenite will be lowered and, as a result, the temperature at which cementite precipitates is also lowered. The slow cooling of the steel is accompanied by constant precipitation of cementite until a temperature of 1335°F. is reached, where the remaining austenite contains 0.85 per cent carbon.

The solubility of cementite in austenite at various temperatures is shown by the line *ES*, which is known as the *cementite solubility curve*, since it is along this curve that cementite is precipitated.

It has been seen that, regardless of the original carbon content of the austenite, when a temperature of 1335°F. is reached the carbon content has become 0.85 per cent. At this temperature, austenite of this composition recrystallizes to form a complex mixture of ferrite and cementite. This point, known as the *eutectoid*, is represented on the diagram by point *S*, the intersection of the ferrite solubility curve *GS* and the cementite solubility curve *ES*.

Phase Regions.—With the phases properly identified, the diagram may now be subdivided into regions based upon the carbon content:

1. 0 per cent C—ferrite.
2. 0 to 0.85 per cent C—hypoeutectoid steels.
3. 0.85 per cent C (approximately)—eutectoid steels.
4. 0.85 to 1.7 per cent C—hypereutectoid steels.
5. 1.7 to 4.3 per cent C—hypoeutectic cast irons.
6. 4.3 per cent C—eutectic cast iron (ledeburite).
7. 4.3 to 6.67 per cent C—hypereutectic cast irons.
8. 6.67 per cent C—cementite.

THE CRITICAL RANGE

From the previous study of heating and cooling curves and of constitution diagrams, it has been seen that where there is a phase change there will be an evolution of heat on cooling and an absorption of heat on heating. By plotting these arrests, the constitution diagrams have been formed. In other words, the arrests indicate the temperatures at which the phase changes take place. These points are called the *critical points.*

The critical points, halts, or arrests in heating or cooling, are designated by the letter *A*, standing for the French word, *arrêt.* In order that the arrests on heating and cooling may be differentiated, the letter *r*, standing for *refroidissement,* is used for cooling and the letter *c*, standing for *chauffage,* is used for heating. Thus, the arrest on heating is called the *Ac* point and on cooling the *Ar* point. Where there are several arrests, as in the iron-carbon system, they are distinguished from one another by numbers following the letters, being numbered in the order of occurrence on heating.

The constitutional diagram shows the position of the critical points under conditions of extremely slow heating or cooling and does not indicate their position when any other rate is employed. It is found that, when rates different from those specified under the conditions of the diagram are employed, the critical points do not occur at the same temperature on heating or cooling. This *lag* in the attainment of equilibrium conditions is termed hysteresis, which implies a resistance of certain bodies to undergo a certain transformation when this transformation is due. Therefore, the *Ac* point occurs at a temperature somewhat higher than would be expected. Similarly, the *Ar* point is somewhat lower. This difference between the heating and cooling criticals varies with the rate of heating or cooling. In other words, the faster the heating the higher will be the *Ac* point, and the faster the cooling the lower will be the *Ar* point. The faster the heating and cooling rate the greater will be the gap between the *Ac* and *Ar* points of the reversible point *A*. In cooling a piece of steel, it is of utmost interest and importance to note that the rate may be so rapid—for example, in quenching the steel in water —as to prevent the full transformation at *Ar*. The reason for this is that the reaction rate decreases with decrease in temperature. In other words, as will be brought out more fully in later discussion, time is a very important factor in bringing about complete transformation at the *Ar* critical.

The various arrests, in the absence of hysteresis and in the order of their occurrence, are given in Table 2-VI.

TABLE 2-VI

Arrest	Temperature, °F.	Significance
A_0	415	A magnetic change in cementite. Probably not a phase change
A_1	1335	Eutectoid transformation
A_2	1414	A change in the magnetic and certain other properties of alpha iron. No phase change
A_3	1335 to 1663	On cooling, the beginning of the precipitation of ferrite from austenite. On heating, the end of the solution of ferrite into austenite
A_4	2557 (in pure iron)	Transformation from gamma to delta iron on heating and the reverse effect on cooling

Since all heat-treating operations are carried out on the material in the solid condition, it becomes very important that consideration be given to those portions of the constitutional diagram that show the reactions occurring in the solid steel as it changes temperature. The portions of particular importance include the lines *GS*, *ES*, *PSK*, *GP*, and *PQ*. The first three of these lines, which are of particular importance, are known as the *thermal critical lines*, and the temperature range that they bound is called the *critical range.* These thermal critical lines are usually referred to as follows: line *PS* as the A_1, line *GS* as the A_3, line *ES* as the A_{cm} and line *SK* as the $A_{1.3}$. In addition to these lines, there is another line, known as the A_2, which occurs at a temperature of 1420°F. (770°C.). This line apparently deviates very little from the temperature at which it is shown as the carbon in the steel is varied. It represents the temperature above which alpha iron becomes nonmagnetic. There is another line, the A_0 line, which is similar to the A_2 and occurs at a temperature of 410°F. (210°C.). It represents the temperature above which cementite is nonmagnetic. Other than these effects, the two lines A_2 and A_0 have but slight significance to the steel treater.

In the upper left-hand corner of the iron-carbon diagram lies the area *ABJN*, which has been disregarded thus far in the development of the equilibrium diagram. It is within this area that delta iron, which is identical in structure with alpha iron, is stable. In ordinary heat-treating operations delta iron and the A_4 thermal critical point play a very negligible part; however, in certain alloy steels they become important.

Table 3-VI[1] illustrates the hysteresis effect on hypoeutectoid and eutectoid carbon steels. It shows the approximate critical temperatures of these steels under conditions of heating and cooling with a rate of cooling about the same as furnace-cooling for commercial annealing. This rate is dependent upon several important variables, such as the mass of the piece, the surface exposed, furnace construction, ratio of furnace size to size and surface of piece.

S.A.E. STEEL SPECIFICATIONS

In the Table 3-VI you will notice that certain S.A.E. (Society of Automotive Engineers) numbers have been listed, with the specification ranges for carbon, manganese, phosphorus, and sulfur in the following columns. These specification numbers are frequently found in literature, particularly where specifications are being quoted for ferrous materials that are necessary

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[1] MORRIS, SERGESON, and GABLE, Approximate Critical Temperatures for S.A.E. Steels, *Metal Progress*, **28** (No. 2), 40*a* (1935); GILLIGAN, F. P., New S.A.E. Steels, *Metal Progress*, **27** (No. 4), (1935).

for the production of automotive apparatus. In many cases, however, special requirements are made by the purchaser for a much narrower range of specifications on a particular grade.

A number index system has been devised, the S.A.E. numbers, which are partially descriptive of the quality of the materials covered. The first figure of the number indicates the class to which the steel belongs: thus 1 indicates a carbon steel; 2, a nickel steel; and 3, a nickel-chromium steel. In the case of alloy steels, the second figure generally indicates the predominant alloying element. Usually, the last two or three figures indicate the average carbon content in "points," or hundredths of 1 per cent. Thus 1020 indicates a carbon steel and 0.20 per cent carbon (0.15 to 0.25); 2340 indicates a nickel steel of approximately 3 per cent nickel (3.25 to 3.75) and 0.40 per cent carbon (0.35 to 0.45); and 71360 indicates a tungsten steel of about 13 per cent tungsten (12–15) and 0.60 per cent carbon (0.50 to 0.70). The basic numerals for the various qualities of steels specified are

TABLE 3-VI

S.A.E. number	Carbon range	Manganese range	Phosphorus (maximum)	Sulfur (maximum)	On slow heating			On slow cooling		
					Ac_1	Ac_2	Ac_3	Ar_3	Ar_2	Ar_1
1010	0.05–0.15	0.30–0.60	0.045	0.055	1350	1405	1605	1570	1400	1255
1020	0.15–0.25	0.30–0.60	0.045	0.055	1355	1410	1570	1535	1395	1260
1030	0.25–0.35	0.60–0.90	0.045	0.055	1350	1405	1495	1465	1405	1250
1035	0.30–0.40	0.60–0.90	0.045	0.055	1345	....	1475	1455	1395	1275
1040	0.35–0.45	0.60–0.90	0.045	0.055	1340	....	1455	1415	....	1275
1050	0.45–0.55	0.60–0.90	0.045	0.055	1340	....	1425	1390	....	1275
1060	0.55–0.70	0.60–0.90	0.040	0.055	1340	....	1410	1370	....	1275
1070	0.65–0.80	0.60–0.90	0.040	0.055	1345	....	1370	1340	....	1280
1080	0.75–0.90	0.60–0.90	0.040	0.055	....	....	1360	1285		
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	....	....	1360	1285		
1095	0.90–1.05	0.25–0.50	0.40	0.055	....	....	1360	1290		

Carbon steels	1	Chromium steels	5
Nickel steels	2	Chromium-vanadium steels	6
Nickel-chromium steels	3	Tungsten steels	7
Molybdenum steels	4	Silico-manganese steels	9

A complete list of S.A.E. specifications may be had by referring to the A.S.M. "Metals Handbook," pp. 612–616, 1939.

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CHAPTER VII

THE METALLOGRAPHY OF STEEL CASTINGS AND CAST IRON

SOLIDIFICATION OF IRON-CARBON ALLOYS

The description of the solidification and cooling to room temperature of typical iron-carbon alloys will now be made in detail. This description is based upon the iron-carbon constitution diagram, Fig. 2-VI. The magnetic transformations A_2 in ferrite and the A_0 in cementite will be omitted in the discussion.

Alloys from 0 to 0.55 Per Cent Carbon.—At the liquidus line *AB*, crystals of delta iron start to solidify from the liquid melt. Upon continued cooling, these crystals are able to hold in solid solution an amount of carbon, as indicated by the line *AH*, and are saturated with respect to carbon upon reaching a value of 0.08, which is on the peritectic line *HJB*. The liquid metal by the precipitation of delta iron is also enriched in carbon, following the liquidus from *A* to *B* until it also reaches *HJB*. If the melt is to the left of *J*, the delta iron of composition *H* will undergo a peritectic reaction with liquid of composition *B* that will result in the formation of austenite of composition *J*, leaving the austenite of composition *J* in equilibrium with a reduced amount of delta iron of composition *H* according to the following reaction:

$$\text{Solid } J \rightleftarrows \text{solid } H + \text{liquid } B$$

On further cooling, we have a slight change in the solid solubility of the two solids *H* and *J* as denoted by the line *HN* and *JN*. Assuming, of course, that the cooling in all cases is slow enough to allow for proper diffusion, the final solid will be austenite containing an amount of carbon in solid solution equal to the original composition of the melt.

If the composition of the melt is to the right of *J*, delta iron of composition *H* will react with liquid at *B* to form austenite of composition *J*, having in equilibrium austenite of composition

J, and a reduced amount of liquid melt of composition *B*. On further cooling, the alloy will completely solidify to austenite upon reaching *JE*.

It can be plainly seen from the diagram that, regardless of the composition of the austenite within this carbon range, the subsequent changes will be identical. These changes are illustrated by the following discussion of a 0.20 per cent carbon melt.

Hypoeutectoid Steel of 0.20 Per Cent Carbon.—On the cooling of such a melt, the austenite thus formed makes no physical change until it reaches the ferrite solubility line *GS* or the upper critical A_3. At this critical, ferrite starts to precipitate and continues to do so as the temperature is progressively lowered, the residual austenite all the time becoming richer in carbon, its composition at any particular temperature being denoted by line *GS*. Upon reaching a temperature of 1335°F., enough ferrite has been rejected from solid solution to make the residual austenite eutectoid in composition. The austenite is then transformed at this point to pearlite. Below this point the structure will be ferrite and pearlite. The microstructure, as shown in Fig. 1-VII*A*, shows the predominant free ferrite, the light areas, and the small amount of dark-colored pearlite.

Hypoeutectoid Steels 0.55 to 0.85 Per Cent Carbon.—The solidification of steels of this series is identical with those in that type of diagram in which there is complete solubility in the solid state. Thus, solidification starts on line *BC*, with complete solidification of the austenite upon reaching line *JE*, the exact temperature depending upon the original composition. The reactions taking place on further cooling to room temperature are identical with those encountered in cooling the 0.20 carbon steel with the structure at room temperature, in all cases being free ferrite and pearlite. Figure 1-VII*B* shows the microstructure of a 0.40 per cent carbon steel and Fig. 1-VII*C* shows that of a 0.60 per cent carbon steel.

Eutectoid Steel, 0.85 Per Cent Carbon (Approximate).—A melt of eutectoid composition will start to solidify at line *BC* and, with continued cooling, will become solid austenite of eutectoid composition upon reaching line *JE*. Upon subsequent cooling from this point there will be no further physical change until a temperature of 1335°F. is reached. At this temperature there is complete transformation of the austenite to lamellar

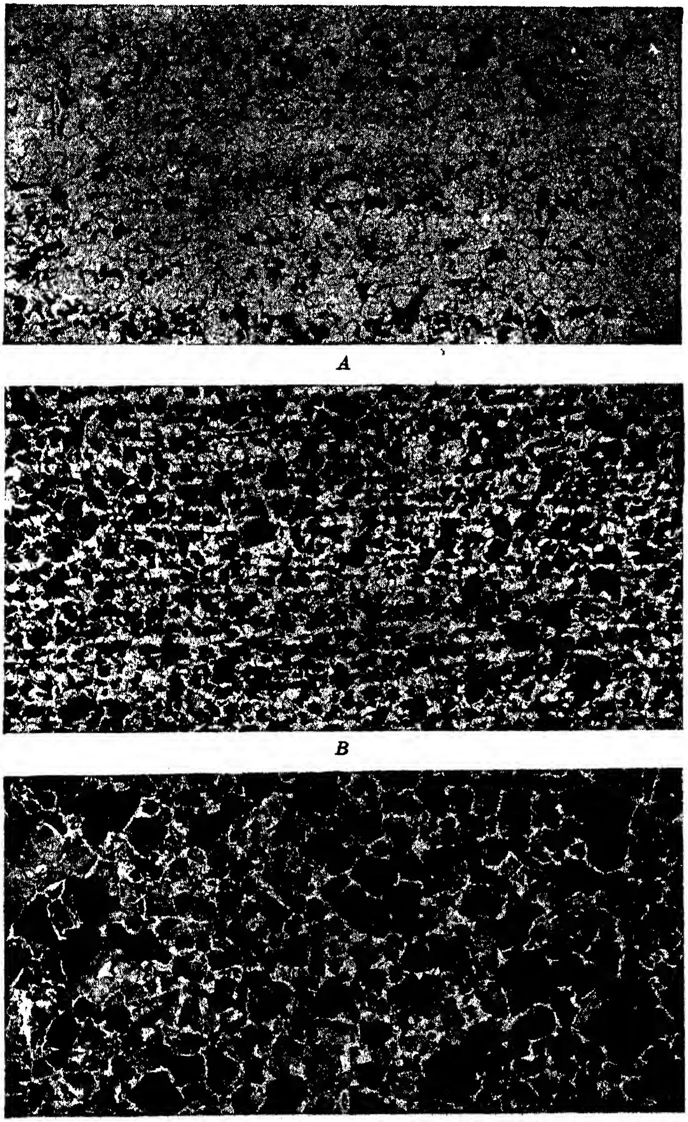

FIG. 1-VII.—*A*, Structure of a 1020 carbon steel after annealing from 1450°F. × 100. Nital etched. *B*, Structure of a 1040 carbon steel after annealing from 1450°F. × 100. Nital etched. *C*, Structure of a 1060 carbon steel after annealing from 1450°F. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

pearlite. There will be no further changes in structure below this point. Figure 2-VII*D* is a photomicrograph of eutectoid steel taken at a magnification of 100 diameters. Photographs at higher magnifications will be shown in the subsequent chapters.

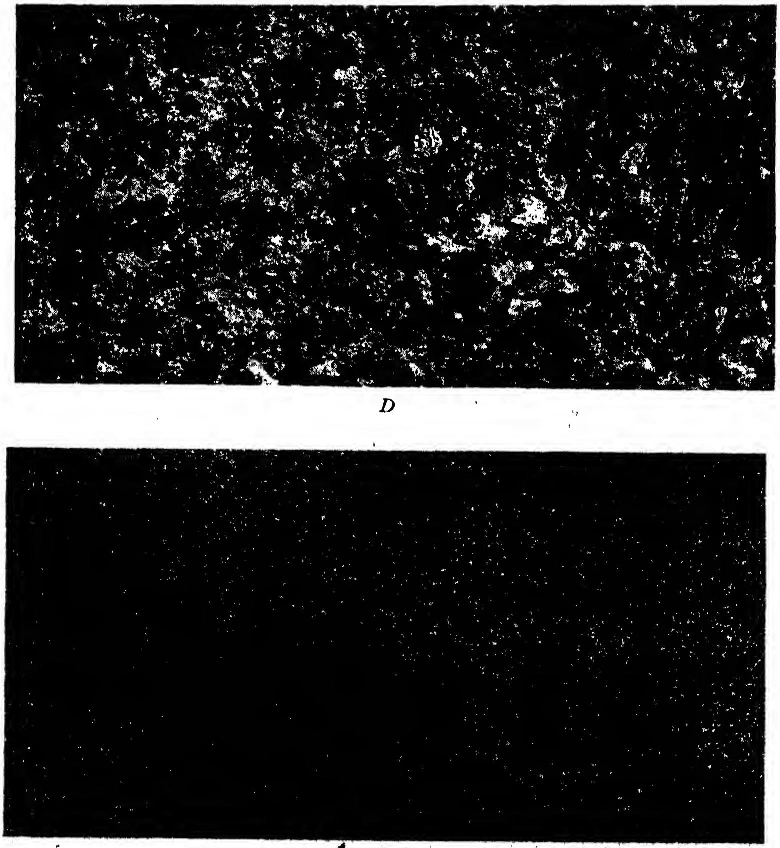

FIG. 2-VII.—*D*, Structure of a 1085 carbon steel after annealing from 1450°F. × 100. Nital etched. *E*, Structure of a 10120 carbon steel as spheroidized, annealed. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

Hypereutectoid Steels, 0.85 to 1.7 Per Cent Carbon.—Solidification of austenite will start at *BC* and, with decreasing temperature, will become completely solidified upon reaching *JE*. Upon continued cooling, austenite of this composition will reach the A_{cm} upper critical, which is the cementite solubility line

SE, and cementite will start to be rejected. On further cooling, cementite rejection is continued while the residual austenite, which is becoming leaner in carbon, follows, in composition, along the A_{cm} line and approaches the eutectoid. At a temperature of 1335°F., cementite rejection is completed and the residual austenite, having reached eutectoid composition, transforms into pearlite. The structural composition below this temperature then is pearlite and free cementite. Figure 2-VII*E*, a photomicrograph of a 1.20 per cent carbon steel, shows the typical structure of this steel after slow cooling (commercial treatment). The light spheroidal areas in this case are free cementite, while the surrounding material is ferrite. This apparent discrepancy in structure will be discussed later in the text.

Hypoeutectic Cast Irons, 1.7 to 4.3 Per Cent Carbon.—Solidification of austenite will start at *BC* and, with continued cooling, the precipitating austenite will follow along line *JE* and approach the point *E*, which denotes the maximum solubility point of carbon in gamma iron. The liquid during this cooling is approaching the eutectic composition *C*. At 2065°F. solidification is complete; the solid consists of free saturated austenite and the eutectic known as *ledeburite*. Ledeburite is a mechanical mixture of approximately 52 per cent of cementite[1] and 48 per cent of saturated austenite.[2] Upon further cooling, the cementite of the eutectic remains unchanged, but the free saturated austenite at *E* and the saturated austenite of the eutectic start to reject cementite from solution. The austenite follows the A_{cm}, or cementite, solubility line and approaches eutectoid composition. At 1335°F. the residual austenite reaches eutectoid composition and transforms to pearlite. The structure below this temperature then consists of free cementite and pearlite.

Hypereutectic Cast Irons, More than 4.3 Per Cent Carbon.—In this group, cementite solidification will start at *CD* and, with continued cooling, there will be continuous precipitation of cementite, with the remaining liquid approaching eutectic composition at *C*. At 2065°F. solidification is complete; the solid consists of free cementite and the eutectic. With continued cooling, the saturated austenite of the eutectic will start to reject

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[1] This may be referred to as eutectic cementite.

[2] This may be referred to as eutectic austenite.

cementite from solution. The residual austenite during the rejection of cementite follows the A_{cm} line and approaches eutectoid composition. At 1335°F. the residual austenite reaches eutectoid composition and transforms to pearlite. The structure below this temperature then consists of free cementite and pearlite.

All the transformations discussed take place in the identical fashion on heating, only in the reverse order.

STRUCTURAL CONSTITUTION OF IRON-CARBON ALLOYS

It will be noticed, upon reviewing the photomicrographs of the iron-carbon alloys, that as the carbon content increases from 0.00 per cent to approximately 0.85 per cent, the amount of free ferrite decreases from what we shall say is 100 per cent at 0.00 per cent carbon to 0.00 per cent at 0.85 per cent carbon. At approximately 0.85 per cent there is no free ferrite but 100 per cent pearlite. The pearlite has increased from 0.00 per cent at 0.00 per cent carbon to 100 per cent at 0.85 per cent carbon. Then, as the carbon content increases from 0.85 per cent, the amount of free cementite present increases from 0.00 per cent at that point to 100 per cent at 6.67 per cent carbon. The pearlite decreases from 100 per cent at 0.85 per cent carbon to 0.00 per cent at 6.67 per cent carbon.

Frequently it becomes important to know the quantities of the structural constituents present in some alloy. This can be accomplished by experience and study of the microstructure, but it can also be calculated with the assistance of the constitution diagram in accordance with the lever-system principle. Typical examples of the use of the lever reaction in determining the amounts of phases in given iron-carbon alloys are as follows: (Assume that the solubility of carbon in "pure iron" at room temperature is zero).

Calculate the amount of free ferrite and pearlite in a S.A.E. 1020 steel at room temperature.

$$\underbrace{0 \qquad 0.20}_{x}\underbrace{\qquad\qquad 0.85}_{y}$$

x = amount of pearlite

y = amount of ferrite

$$x = \frac{0.20}{0.85} \times 100 = 24 \text{ per cent pearlite}$$

$$y = \frac{0.65}{0.85} \times 100 = 76 \text{ per cent ferrite}$$

Calculate the amount of free cementite and pearlite in a S.A.E. 10120 steel at room temperature.

$$\underbrace{0.85 \qquad 1.20}_{x}\underbrace{\qquad\qquad\quad 6.67}_{y}$$

x = amount of cementite
y = amount of pearlite

$$x = \frac{1.20 - 0.85}{6.67 - 0.85} \times 100 = \frac{0.35}{5.82} \times 100 = 6 \text{ per cent cementite}$$

$$y = \frac{6.67 - 1.20}{6.67 - 0.85} \times 100 = \frac{5.47}{5.82} \times 100 = 94 \text{ per cent pearlite}$$

Calculate the amount of ferrite and cementite in a eutectoid steel at room temperature.

$$\underbrace{0 \qquad 0.85}_{x}\underbrace{\qquad\qquad\quad 6.67}_{y}$$

x = amount of cementite
y = amount of ferrite

$$x = \frac{0.85}{6.67} \times 100 = 12.7 \text{ per cent cementite.}$$

$$y = \frac{6.67 - 0.85}{5.67} \times 100 = \frac{5.32}{6.67} \times 100 = 87.2 \text{ per cent ferrite}$$

This last calculation proves mathematically the ratio of ferrite to cementite, mentioned earlier, to be approximately 7 to 1.

By making the proper calculations to determine the amounts of the phases present for a given iron-carbon alloy from its melting temperature to room temperature, a diagram can be drawn that will show the structural changes and the amounts of the various phases present. Figure 3-VII shows such a diagram for a 1020 carbon steel as it slowly cools from the molten condition to room temperature. Figure 4-VII shows a diagram for a 10120 carbon steel.

PHYSICAL PROPERTIES OF THE CONSTITUENTS

In Table 1-VII the physical properties of ferrite, cementite, and pearlite, under normal conditions, are shown.

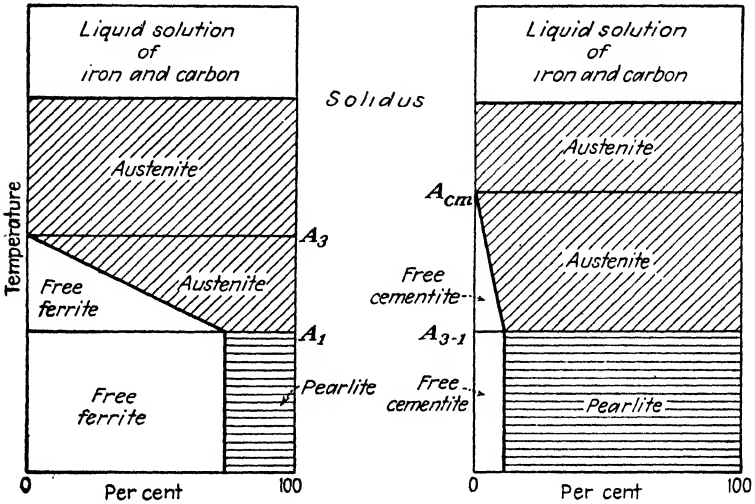

FIG. 3-VII. FIG. 4-VII.

FIG. 3-VII and FIG. 4-VIII.—Diagrams showing structural changes and per cent phase present at any given temperature for a 1020 steel (Fig. 3-VII) and for a 10120 steel (Fig. 4-VII) as it slowly cools from the molten condition to room temperature.

When the physical properties of these three constituents are known it should be possible to estimate with some degree of accuracy the physical properties of any steel of known structural composition, assuming that the steel is in the pearlitic condition or that the conditions of cooling have been normally slow. This

TABLE 1-VII

Iron-carbon alloy	Specific gravity	Tensile strength, p.s.i.	Elongation in 2 in., per cent	Brinell hardness number
Ferrite	7.86	38,000–40,000	40	90
Pearlite	7.846	120,000 ±	15	240
Cementite	7.66	5,000	None	550 ±

possibility of foretelling the physical properties is based on the assumption that the constituents present impart their physical properties to the steel in the proportion to which they are present. From this consideration, then, it would seem evident that a eutectoid steel must possess maximum tensile strength and that the presence of either free ferrite in hypoeutectoid steels or free cementite in hypereutectoid steels would have a weakening influence, owing to the relatively low strength of both free ferrite and cementite. The diagram in Fig. 5-VII illustrates the effects of carbon content with respect to the tensile strength, hardness, and ductility.

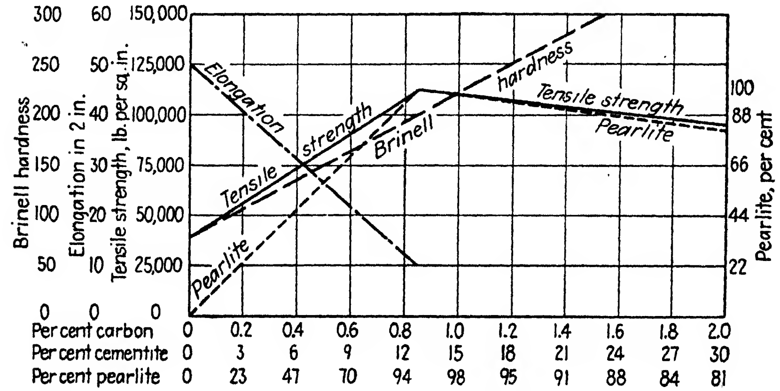

FIG. 5-VII.—Diagram showing the relations between tensile strength, hardness, and ductility, and the carbon content. (*After A. Sauveur.*)

Suggested Questions for Study and Class Discussion

1. Fully define the following, limit the temperature and composition range of each, and describe its structure.

Delta iron
Gamma iron, or austenite
Alpha iron, or ferrite
Iron carbide, or cementite
Pearlite

2. What is meant by the critical-transformation points? Do these points occur at the same temperature on heating and cooling in commercial operations?

3. Describe the structural changes on cooling from above the point of solidification to room temperature and the structure at room temperature of each of the following alloys: S.A.E. 1010, S.A.E. 1040, S.A.E. 1090, 1.15 per cent carbon steel.

4. What structural changes take place on heating the following?

S.A.E. 1010 to 1250°F.; to 1750°F.
S.A.E. 1040 to 1550°F.
1.15 per cent carbon steel to 1650°F.

5. How much pearlite and how much ferrite will be present in an S.A.E. 1010 steel at room temperature? How much pearlite and how much free cementite will be present in a 1.15 per cent carbon steel at room temperature? How much ferrite and how much cementite will there be in a commercial S.A.E. 1090 steel?

6. Under normal conditions, compare the following steels, S.A.E. 1010, S.A.E. 1040, 1.15 per cent carbon, with respect to tensile strength, hardness, and ductility.

CHAPTER VIII

THE MECHANICAL TREATMENT OF STEEL

The primary purpose of mechanical treatment is, as the term implies, the reduction by mechanical means of a given section of steel into a useful product. Mechanical shaping has the added function of improving the quality of the steel by increasing the homogeneity of the reduced section, closing cavities, and refining the crystalline structure. By making the proper selection of the temperature at which the working is carried on and the application of the work through the rolls, dies, etc., the physical properties—*viz.*, strength, ductility, and hardness—are all definitely affected. The strength is always increased, the hardness may or may not be increased, while the ductility may either be increased or decreased, depending upon the working conditions. The amount of change in each of these properties for a given steel is affected by the amount of work performed and by the temperature at which the work is carried on.

CAST STEEL

A study of cast steel is of particular importance because it is a valuable finished material and is also the intermediate product for all wrought steels. Obviously any variations in its structure and properties may, when it is subjected to hot or cold work, be reflected in the structure and properties of the sections made from it.

Distinction between Cast Steel and Steel Castings.—It is important, in discussing the subject, to differentiate between cast steel and steel castings.[1]

Cast steel is defined as a commercial iron-carbon alloy containing less than 1.7 per cent of carbon, melted and poured into a mold of any material, size, or shape. Cast steel may be used commer-

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[1] Sisco, F. T., "The Alloys of Iron and Carbon," Vol. II, Properties, McGraw-Hill Book Company, Inc., New York, 1937.

cially in the form in which it is cast or it may be worked hot or cold into an industrial product.

Steel castings are defined as cast steel that has been poured into molds of the shape desired for the finished product and are so used industrially without any mechanical treatment.

Solidification.—Prior to the study of the treatment of either cast steel or steel castings, consideration should be given to the solidification and cooling phenomena, so that subsequent treatment can be effected according to the condition of the structure. The solidification of a steel casting cooled slowly from the molten condition to room temperature is as follows: Initial solidification of the molten metal starts immediately upon contact with the comparatively cold walls of the mold, where a relatively thin skin of small equiaxed crystals is formed. As the cooling rate of the steel decreases, a second zone of crystallization will start and proceed by growth inwardly at right angles to the mold walls. This crystallization occurs selectively, that is, those elements that raise the melting point of the steel tend, in general, to concentrate in the first parts of the crystals formed, while those elements that lower the melting point will be forced into the regions that crystallize last. Thus, the first crystals in the casting will be relatively pure iron, while the last material will be richer in carbon and impurities. In castings, the growing primary crystals will quickly interfere with each other in all directions except vertically from the wall and, as a result, usually take the form of columnar crystals, or dendrites, extending at an angle of 90 deg. from all walls of the mold. Radial dendritic crystallization is shown in Fig. 1-VIII. While these directional crystals are growing, the liquid in the interior is cooling and finally reaches the solidification temperature, where it also crystallizes but with no directional tendencies.

Since the solidification in the iron-carbon alloys is selective in nature, *i.e.*, the crystallizing solid varies continuously in composition from initial solidification to final, the first portion of a dendrite will differ chemically from parts formed at a later time. The major portion of a dendrite will differ chemically from the later-forming secondary axes, and it will also differ from the filled-in material. Such a condition of heterogeneity is termed *dendritic segregation.* Observation of this dendritic structure is made possible through proper etching. It must be noted, however, that the possibility of revealing a dendritic structure through etching treatment does not depend upon the existence of dendrites but upon the existence of dendritic segregation. This is due to the fact that etching methods depend for their action on differences in composition.

The number, size, and form of these crystals are influenced by a variety of factors, such as the composition of the steel, the initial temperature of the liquid steel, the rate of solidification, the size of the casting, and, apparently, of the type of dispersion of nonmetallic matter in the steel.

Fig. 1-VIII.—Dendritic crystallization in steel. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

Recrystallization.—In addition to the above conditions concerning the primary crystallization of steels, account must also be made of the conditions involved in the allotropic transformation of gamma to alpha iron and the reverse, and the resultant grain size. This process, whereby a new crystalline form or even a new set of crystals of like form is produced from preexisting crystalline material, is known as *recrystallization.*[1]

Thus, if a carbon steel, either hypoeutectoid or hypereutectoid in character, of average austentic grain size, is slowly cooled, ferrite or cementite is rejected largely at the grain boundaries of the austenite as it passes through the critical range; after which the remainder of the austenite transforms to pearlite. This new structure has its own specific grain size and shape properties; however, it may be true that some of the dimensional characteristics of the austenite may be inherited by the new structure. This recrystallization apparently takes place in a fashion similar to primary crystallization, in that it is nuclear in origin, *i.e.*, the new crystal structure of one modification does not change into one of the other as a whole, but rather the transformation proceeds from numerous nuclei within the old crystals. Owing to the fact that the rejection of ferrite or cementite starts at nuclei in the austenitic grain boundaries as it passes through the critical range, a pure grain-size effect is obtained only in eutectoid steels. The rejected grain-boundary material in the hypo- and hypereutectoid steels and the rate of cooling through the critical,[2] which affects the amount of boundary ferrite or cementite, has its effect on the grain characteristics of the steel.

In addition to recrystallization on cooling, we must concern ourselves also with the recrystallization on heating. When a piece of carbon or low alloy steel is heated above the critical range of the material, a reaction will occur that causes a transformation to the new crystalline phase, austenite. Consider now the heating of a hypoeutectoid steel to above its critical range. As the steel is slowly heated to just above the Ac_1 point, small grains of austenite will be formed within the pearlite regions.[3] As the temperature is raised, each little grain of austenite will start to grow until a considerable portion of the pearlite has

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[1] This refers to recrystallization produced by thermal treatment and not to recrystallization resulting from cold work. The two phenomena, however, are similar in many respects.

[2] McBride, Herty, and Mehl, The Effect of Deoxidation on the Rate of Formation of Ferrite in Commercial Steels, *Trans. A.S.M.*, **24**, 281–312 (1936).

[3] Grossmann, M. A., On Grain Size and Grain Growth, *Trans. A.S.S.T.*, **21** (No. 12), 1079–1111 (1933).

transformed, at which time the austenite will encroach into the ferrite grains. In the ideal case, each of these original crystals would grow until it reached its similarly growing neighbor and there would be just as many austenite grains as there were original nuclei.

However, some of the grains of austenite grow more rapidly than others and will absorb the smaller grains to form large ones. This condition probably occurs frequently in some steels and rarely in others.

Similar conditions of refinement, as described for the hypoeutectoid steels, will take place during the heating of both eutectoid and hypereutectoid steels. It must be emphasized, however, that if the steel is other than eutectoid in character (containing free ferrite or free cementite) and has been heated to just above the lower critical, there will not be complete refinement. The excess ferrite and cementite, in this case, will not be changed and the steel will retain practically its original grain size. Complete refinement of the structure can be effected only by heating to a temperature slightly above the upper critical.

The resultant austenitic grain size on heating a carbon steel above the Ac_3 or Ac_{cm} temperature is largely dependent upon the number of nuclei present and, according to Herty and his coworkers,[1] upon the rate of heating through the transformation range in the case of those steels in which there has been no aluminum addition. As the rate is increased, the initial austenitic grain size becomes larger and, after a period of about 20 min. from the time the operating temperature is reached, grain growth is practically complete. In the case of aluminum-killed steels, however, the rate of heating has very little influence; all rates yield small grains.

Phase Precipitation.—Now, as to the mode of occurrence of the precipitating phase with respect to the rate of cooling, let us take a grain of austenite in a cast hypoeutectoid steel and see what happens during the cooling through the critical range. According to findings of a number of investigators, this grain of austenite in passing through the critical range rejects first proeutectoid

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[1] HERTY, MCBRIDE, and HOUGH, The Effect of Deoxidation on Grain Size and Grain Growth in Plain Carbon Steels, *Cooperative Bulletin* 65, Carnegie Institute of Technology and Mining and the Metallurgical Advisory Boards, 1934.

ferrite to its boundary, thus outlining the original austenitic grains, with the remaining austenite being converted into pearlite. If the cooling is quite rapid, as is the case in foundry practice, only part of the ferrite is ejected to the boundary with the remaining austenite, now containing ferrite in excess of eutectoid ratio, forming a finer lamellar pearlite. Microscopic examination would then show a grain with ferrite boundaries enclosing a pearlitic inner structure. With somewhat slower cooling, and in the presence of the ferrite already at the boundaries, there will be more complete rejection of the ferrite. This ferrite either joins the ferrite at the boundaries or separates preferentially along certain crystallographic planes within the grains.

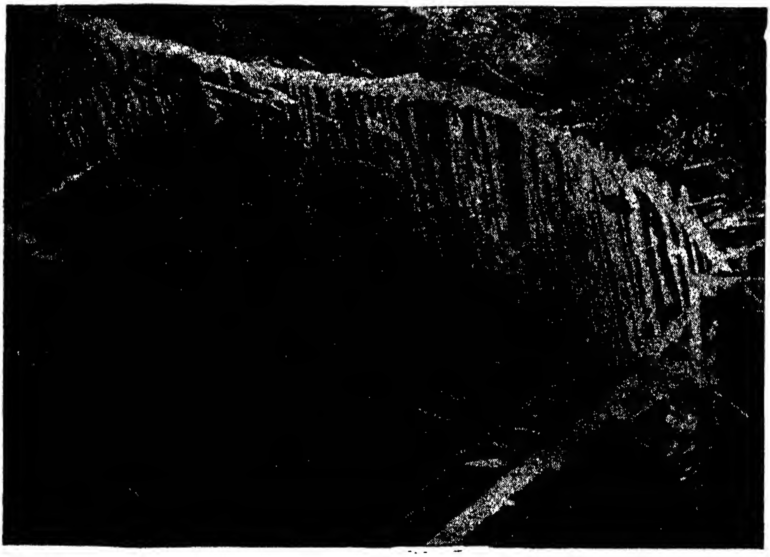


FIG. 2-VIII.—Cast steel, 0.40 per cent carbon. × 100. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

The result of the so-called *intracrystalline* precipitation of the ferrite along certain crystallographic planes produces a characteristic microstructure known as the *cleavage* or *Widmanstätten* pattern. Figures 2-VIII and 3-VIII show typical *Widmanstätten* structures in which varying cooling rates have produced a varying amount of intracrystalline ferrite.

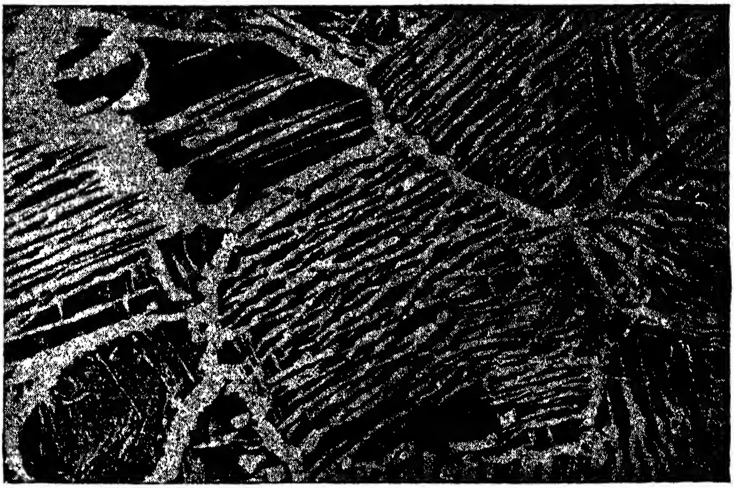


FIG. 3-VIII.—Cast steel, 0.50 per cent carbon. × 100. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

FIG. 4-VIII.—Cast steel, 0.55 per cent carbon. Section parallel to the surface of a cube. × 30. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

It can be seen that the intracrystalline ferrite has had a preferred precipitation along certain planes, which planes lie at 60-deg. angles. This peculiarity is due to the fact that in all face-centered cubic structures (austenite) the preferred precipitation occurs on the so-called *octahedral* planes in the lattice structure that lie at the angle observed. The pattern in Fig. 3-

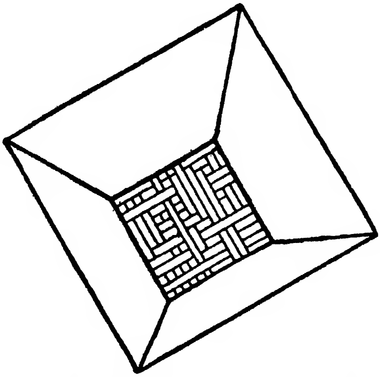


FIG. 5-VIII.—Section parallel to the surface of a cube. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

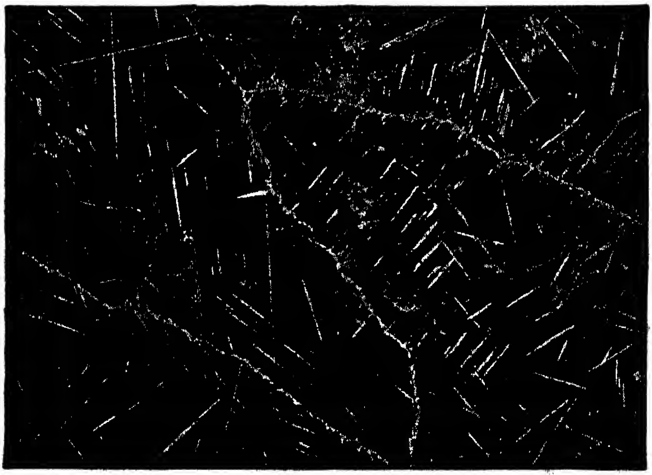


FIG. 6-VIII.—Hypereutectoid cast steel. Free cementite rejected partly to the boundaries and partly between cleavage planes. × 114. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

VIII does not indicate perfect angularity throughout, owing to the fact that the grains are not oriented perfectly with respect to each other in the material. This manner of precipitation, however, is not an invariable occurrence, for other planes than the octahedral planes may, at times, be involved, as is illustrated in Figs. 4-VIII and 5-VIII.

A similar action takes place in hypereutectoid steels, although it is somewhat more sluggish than in hypoeutectoid steels, with cementite being the precipitating phase (Fig. 6-VIII).

INGOT STRUCTURE

The ingot into which steel is cast represents a considerably larger steel casting than has been considered in our previous discussion. Therefore, it will have similar inherent properties: *i.e.*, being dendritic, having platelets or planes of cleavage, and having a coarse crystalline structure. In this condition the steel lacks ductility and strength and is heavily stressed. Therefore, for improving the quality of the steel it must be mechanically worked.

MECHANICAL WORK

In considering the mechanical working of steel, we shall first make a relative distinction between hot and cold working. We know that, with respect to temperature and composition, as soon as steels have solidified they may be worked. This work may be started and continued from the temperature of solidification to room temperature, which covers a range of some 2500°F. We may now make a sharp differentiation and say that all work performed at high temperature, or temperatures exceeding the A_3 and $A_{3.1}$ criticals, will be called hot work and that cold work is that work performed at low temperatures or below the critical range, namely A_1. Hot working, then, implies working the steel while it is in the austenitic condition. At these temperatures the steel will start to recrystallize as soon as the work has ceased, provided, of course, that the finishing temperature is above the critical. Cold work, on the other hand, implies the working of an aggregate that will not recrystallize upon cessation of the work—at least, as far as the distorted pearlite is concerned.

Hot Work.—In following the effect of hot work on steel we will assume that the ingot has been cooled to a temperature low enough so that reheating to above the critical is necessary before hot working is started. As the steel is reheated to higher and higher temperatures and is transformed into the austenitic condition, it becomes weaker and weaker but more plastic, and the grains will grow in size.

It is important to note that this condition of increased grain growth on heating beyond the Ac_3 point is a condition that exists only in the so-called *true coarse-grained* steels. The effects of temperature and time on the growth characteristics of given steels will be discussed in Chap. XIII, but briefly the conditions are as follows: If the steel-melting practice has been so modified as to control the grain size (growth), it will be found that for some steels there is a very slight increase in grain size until we pass a certain definite temperature, which may be as much as 500°F. above the Ac_3 critical. The position of this grain-growth range is a function of the deoxidation practice and is affected by the condition of the liquid steel, the kind and amount of deoxidizers added, and the temperature of the metal.

From Fig. 7-VIII the influence of mechanical work on the structure and grain size of one particular steel will be studied. The results with steels of other compositions and grain-coarsening characteristics will deviate somewhat from the trends indicated. The critical range both on heating and cooling is represented by a double line. This, we know, is incorrect, as the critical range on heating is somewhat higher than the equilibrium temperature and that on cooling somewhat lower; but such a representation with only two critical points is preferable for the sake of simplifying the discussion. The solidification range is similarly represented. The widths of the shaded areas represent the proportional grain size with reference to the treatment. The areas that have been reduced to a line denote that the corresponding grain is very small.

In the solidification range at *A*, the steel solidifies with the formation of austenitic grains of a size denoted by the width of the shaded area. On slow cooling from the temperature of solidification to the critical temperature, the only change that occurs is the coarsening of the austenitic grain, as is noted by the slope of the lines bounding the shaded area. While cooling takes place, the austenite changes to pearlite and ferrite or to pearlite and cementite, depending upon the carbon content, the grain size of the new structure being approximately the same as the austenitic grains from which it was formed. Slow cooling from the lower critical to room temperature does not change the structure.

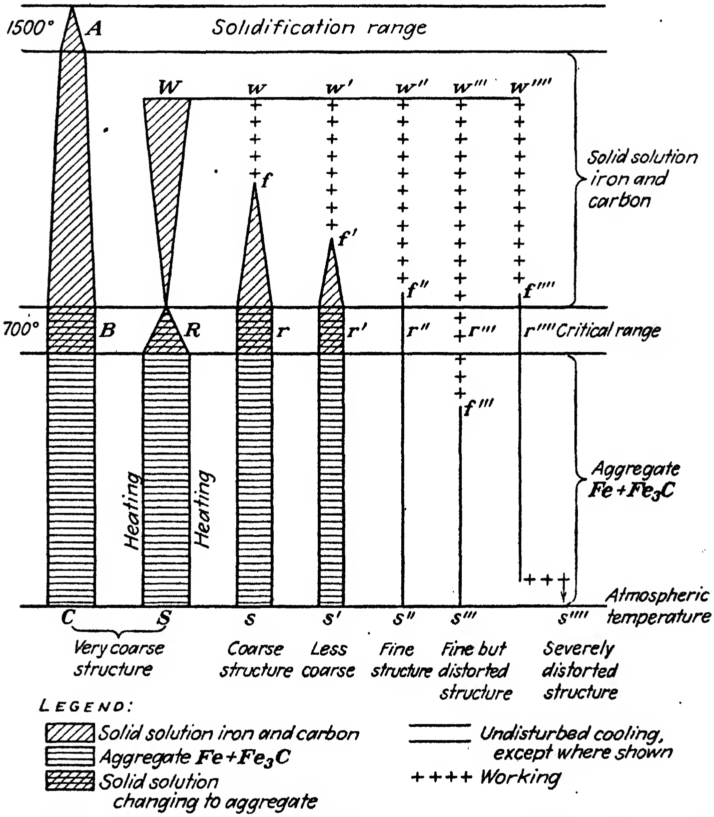


FIG. 7-VIII.—Diagram showing the effects of work on the structure of a particular steel. (*After A. Sauveur.*)

On the ingot's being reheated from *S* to the lower critical, there is no transformation, but on slow heating through the critical range at *R* there is a gradual solution of the aggregate so that, upon its reaching the upper critical, solution is complete, the constituents are in austenitic solid solution, and the coarse grains have been dissolved to form fine grains. Continued slow heating causes grain growth to take place until, when *W* is reached, there may be the grain size corresponding to the width of the area at that point. Upon the application of mechanical pressure at *w*, this crystallization is broken up, the large austenite grains being distorted and divided up into a large number of small grains, each possessing the characteristics of the original grain. Continued work will prevent or retard grain growth. As soon as the work ceases, as is indicated by point *f*, the metal is left to cool and, should the temperature be above the critical—as we can see that it is in this case—grain-growth will immediately start and continue, as from *f* to *r*, until the critical range has been reached, where there is a conversion of the austenite. The final grain size is indicated by the width of the band at *s*. The grain size is now

Fig. 8-VIII.—Structure of a 0.50 per cent carbon steel as cast. × 100. (*From H. M. Boylston, "An Introduction to the Metallurgy of Iron and Steel."*)

somewhat smaller than the original but still fairly coarse. From this we can make the following conclusion regarding hot working of steel. Working above the critical range gives no permanent distortion of the grains. The size of the grains is dependent upon the temperature above the critical at which the work is completed and upon the amount of work performed within this working range.

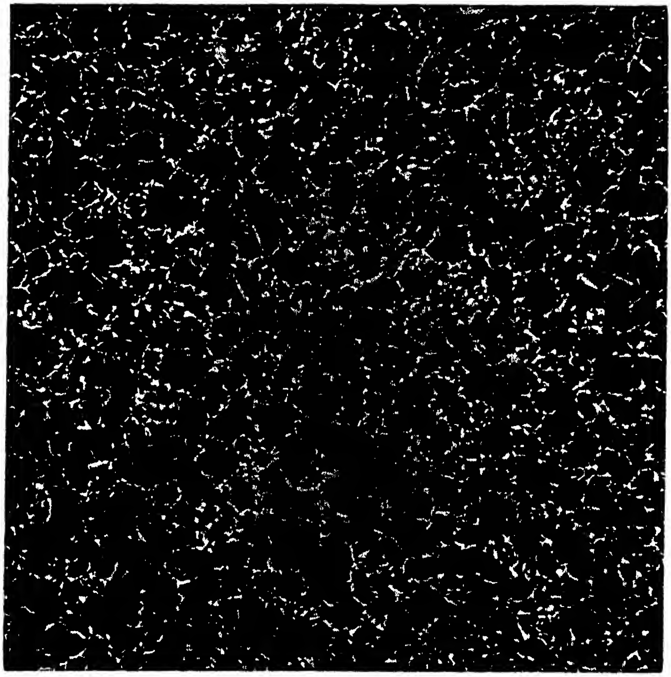


Fig. 9-VIII.—Structure of a 0.50 per cent carbon steel after hot working at low finishing temperature. × 100. (*From H. M. Boylston, "An Introduction to the Metallurgy of Iron and Steel."*)

This conclusion is further shown graphically in the diagram (Fig. 7-VIII). If the work be started at w' and finished at f', the final grain size will be as at s'. Similarly, if the work w'' be continued until the temperature has dropped to f'', at approximately the critical temperature, the final structure will be quite fine. The temperature at which work ceases is known as the *finishing* temperature.

Figures 8-VIII, 9-VIII, and 10-VIII, respectively, show the microstructure of a 0.50 per cent carbon steel as cast and after hot working with a low and a high finishing temperature.

Hot-working Temperature Range.—It has been found that where steel is heated above a range of about 700° F. (370° C.) the strength decreases and the plasticity increases until a point is reached where the lowest melting constituent of the steel starts to melt. At this temperature, which is dependent upon the composition of the steel, the strength is greatly decreased and the metal loses its plasticity. Application of mechanical work at this point will cause the metal to crumble rather than flow.

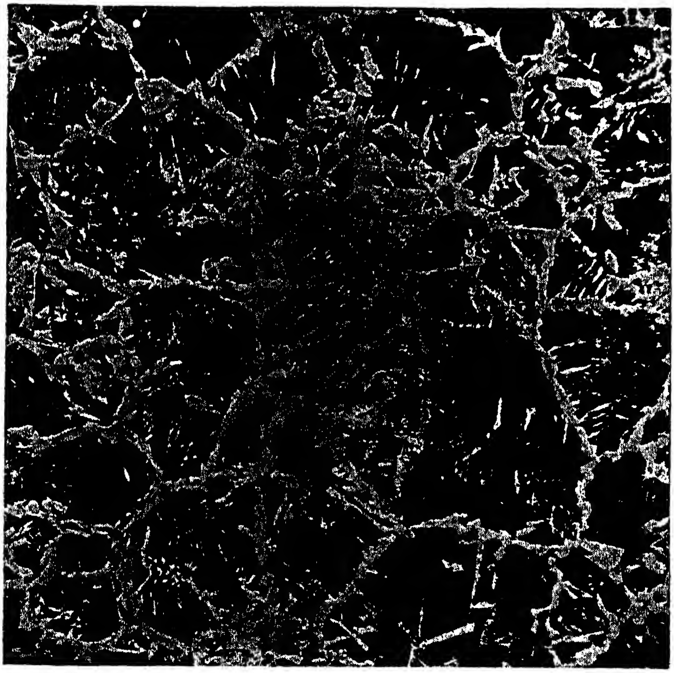


FIG. 10-VIII.—Structure of a 0.50 per cent carbon steel after finishing at high temperature. × 100. (*From H. M. Boylston, "An Introduction to the Metallurgy of Iron and Steel."*)

The top of the hot-working range is that point of safety below the melting point of the lowest melting constituent. The lower point of the range is that point where the strength increases and the plasticity decreases to the point where excessive pressures are required to cause deformation. The usual working temperature is generally some balance between these two extremes.

The actual finishing temperature is determined by the mass, shape, process, and other considerations. It is generally conceded that the lower the finishing temperature, the finer will be the resultant structure and the more reactant the said structure will be toward heat-treatment.

From the previous study of solidification of steel ingots it was learned that the final structure is made up of a mass of grains of varying sizes, each grain being chemically heterogeneous as the result of its solidification characteristics. In addition to this chemical heterogeneity, there will be segregated materials and blowholes. When the material is subjected to hot working the heterogeneous grains, segregated material, blowholes, etc., are elongated in the direction of the working. Macroetching of a section parallel to the direction of work will show these nonmetallic inclusions as lines or bands. These lines or bands are known as *flow lines* because they denote the direction of flow of the metal.

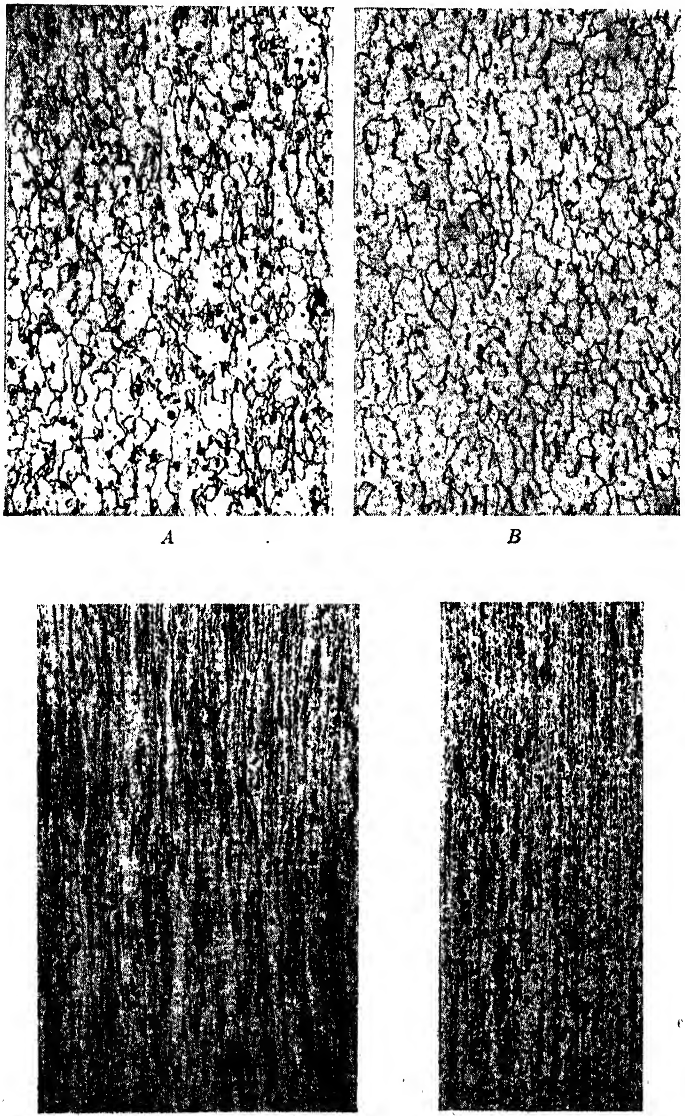
The physical properties, such as tensile strength, ductility, and resistance to shock, are distinctly affected by the directional flow. If the direction of the applied stress is in the direction of greatest elongation, the mechanical properties are improved but, if the stress is applied perpendicularly, the properties are likely to be inferior. Thus, definite consideration must be given to the directional characteristics effected by the forging or rolling practice, as they might affect the service in which the material is placed.

All further hot mechanical working operations will be subject to the factors that have been considered and will have as their purpose (1) improving the quality of the steel by increasing its homogeneity and soundness; (2) forming the metal to the commercial shapes desired; (3) refining the crystal structure; (4) deforming the dendrite segregation, which results in directional properties; and (5) increasing its strength and other physical properties.

Cold Work.—Cold work, as has been noted previously, is mechanical work applied to the metal while its temperature is below the thermal critical range. Incidentally, most of the cold work is done at or about room temperature. It is not at all surprising, in considering the effect of cold work upon the properties of the metal, that the results are found to be substantially different from those resulting from hot work. In the case of hot work, consideration has been given to an entirely austenitic structure, while in that of cold work, consideration will be given to a structure composed of free ferrite and pearlite. This discussion will be limited to the working of the lower carbon range of hypoeutectoid steels.

The effect of cold work is shown graphically in Fig. 7-VIII. If the work is started, as in w''', as hot work and continues through the critical range to a temperature indicated by f''', the structure will be fine, since there has been no opportunity for crystallization of the metal upon passing through the critical range, but it will be distorted as a result of the working in the so-called *cold-working range.* That is, the crystalline elements, which may be ferrite and pearlite, have been elongated in the direction of the working. In the next case, where work has started at w'''', there has been recrystallization and, since the work has been completed close to the critical, the resultant structure will be fine. In this case, the metal was cooled to about room temperature and then was subject to cold work. This work will produce a severely distorted structure, the amount of distortion being dependent upon the amount of work expended.

There is a distinct and important difference between the two operations above. In the first case, we find that upon cessation of the work at f''' the ferrite present in the aggregate recrystallizes according to the same geometric pattern as existed before cold working. The size of the recrystallized ferrite grains is again a function of the temperature and time—in this case, below the critical. The lowest limit of temperature at which this can take place is about 450°C. (842°F.). The distorted pearlite particles remain distorted and can be recrystallized only by heating at or above the critical range. In the latter case, where working was effected below this limiting temperature, there can be no recrystallization without proper heat-treatment. This effect will be discussed later. It might be said that the first graphical representation is rarely, if ever, carried out in practice. This is due to the fact that when those steels other than the aluminum killed and the higher carbon analyses are worked from about 400 to 750°F. (200 to 400°C.) they become quite



A B

C D

Fig. 11-VIII.—*A*, 0.06 per cent carbon steel. As rolled. × 100. Nital etched. *B*, 0.06 per cent carbon steel. Cold-reduced 21 per cent. × 100. Nital etched. *C*, 0.06 per cent carbon steel. Cold-reduced 75 per cent. × 100. Nital etched. *D*, 0.06 per cent carbon steel. Cold-reduced 95 per cent. × 100. Nital etched.

brittle and are almost impossible to work. This temperature range, known as the *blue brittle range,* is avoided in cold-working operations.

The effect of varying amounts of cold work on the structure of a killed low carbon steel is seen in Fig. 11-VIII*A*, *B*, *C*, and *D*.

Cold working deeply affects the physical properties of metals, its action increasing with increasing amounts of cold work. The hardness, tensile strength, and yield point are increased, whereas the ductility properties, such as elongation and reduction of area, are decreased. The effect of cold work is illustrated in Table 1-VIII, which shows the tensile strength, yield strength (0.2 per cent set), and elongation of an 0.08 per cent carbon steel rod cold-worked in increasing total percentages.

TABLE 1-VIII

Physical properties	Cold work, per cent					
	0	33	53	69	76	84
Yield strength, p.s.i.	35,000	77,900	89,500	90,500	93,000	97,500
Tensile strength, p.s.i.	47,200	83,000	93,000	100,000	101,700	106,900
Elongation in 2 in. per cent	27.5	7.0	6.0	4.5	3.5	3.0

In referring to cold-working or cold-drawing operations, it is important to differentiate the processes according to the magnitude of the cold-drawing operations and their resultant effect on the microstructure. In processes such as some wire drawing and strip drawing, where the cold reductions are drastic, the grain structure is definitely elongated in the direction of working and can easily be differentiated from the original hot-rolled structure. In processes involving cold finishing, generally used for bars and shafts, the amount of draft is not sufficient to cause enough structure distortion to be apparent under ordinary microscopic examination.

Hot-rolled versus Cold-drawn Physical Properties.—The physical characteristics of steel are profoundly affected by the cold-working operation. It raises the tensile and yield values, with the yield values increasing proportionately more than the tensile increase. It increases the hardness and decreases the ductility. This general effect is shown in Fig. 12-VIII, which shows the typical stress-strain diagrams for three widely used steels. The specimens used to obtain these results were 1$\frac{1}{16}$-in. rounds hot-rolled, cold-drawn to 1-in. round in regular commercial production.

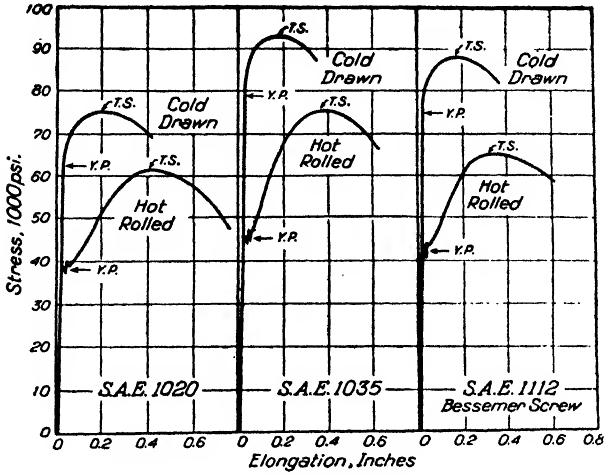


FIG. 12-VIII.—Typical stress-strain diagram of hot-rolled and cold-drawn steels.

A more complete set of physical properties of the three steels in the hot-rolled and cold-drawn condition is shown in Table 2-VIII.[1] The cold-drawn values were obtained on hot-rolled bars cold drawn at $\frac{1}{16}$-in. draft amounting to 11.4 per cent in

TABLE 2-VIII.—REPRESENTATIVE PHYSICAL PROPERTIES, HOT-ROLLED (H.R.) VERSUS COLD-DRAWN (C.D.)

Grade	Size	Yield point	Tensile strength	Elongation, per cent	Reduction per cent	Brinell
S.A.E. 1112	1$\frac{1}{16}$ in. rod, H.R.	40480	64780	35.0	53.0	131
S.A.E. 1112	1 in. rod, C.D.	82460	89700	17.0	43.5	187
S.A.E. 1020	1$\frac{1}{16}$ in. rod, H.R.	38240	59100	37.0	62.0	118
S.A.E. 1020	1 in. rod, C.D.	75120	82750	20.0	51.7	156
S.A.E. 1035	1$\frac{1}{16}$ in. rod, H.R.	45250	77100	35.0	48.0	149
S.A.E. 1035	1 in. rod, C.D.	87140	94240	15.2	41.2	197

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[1] BECK, J. E., "Cold Forming Processes—Drawing Rods and Bars," Jones and Laughlin Steel Corporation, Pittsburgh, 1937.

regular commercial production. A close study of the figures in the table will also show that Bessemer steel has a more pronounced reaction to cold work than has steel made by any other process. This difference in reaction toward cold work is used in many applications where it is necessary to obtain maximum physical properties.

Suggested Questions for Study and Class Discussion

1. Describe the successive crystalline transformations that take place during the solidification of a killed-steel ingot. Will this structure be the same as that found in a cast-steel object, such as a driving wheel center? Why?

2. What effect will such operations as hammering, rolling, or forging have upon the structure and physical characteristics of cast metal on working at temperatures exceeding the A_3 and $A_{3\cdot 1}$ criticals?

3. What are the purposes of hot mechanical working?

4. What effect does cold mechanical work have on the microstructure and physical properties?

5. What differences in the microstructure will be noted if the mechanical work is completed at 1300°F. and allowed to cool to room temperature, or if the working is carried out completely at 200°F.?

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CHAPTER IX

THE HEAT-TREATMENT OF PLAIN CARBON STEELS

In the previous study of the simple binary system, it was found possible, theoretically at least, to predict the microstructure at any particular temperature when the alloy considered was in the state of equilibrium. Under these conditions of equilibrium there is the limiting restriction that sufficient time should be allowed for diffusion, as, for example, the equalization of the differences in cored crystallites. In addition, it was found that, to bring about the desired result of diffusion and equalization, the alloy must be heated to a sufficiently high temperature and maintained at that point for a sufficient length of time. If, on subsequent cooling from this temperature, a very slow rate is employed, there will be normal phase precipitation and the alloy is said to be in the annealed condition. If this rate is increased, as for example, by cooling in air or by more drastic cooling, as by quenching in cold water, the usual precipitation will be retarded or it may even be suspended altogether.

Therefore, according to the rate of cooling employed, the condition of the alloy at high temperatures may be completely transformed as in slow cooling—annealing—or it may be wholly or partially retained by suddenly bringing the alloy to a lower temperature—quenching. By knowledge of this phenomenon we are able to bring about the many remarkable properties of heat-treated steels and numerous other alloys.

ANNEALING

The term "annealing" is rather loosely used to mean any one of a number of treatments that have as their purpose the relief of stresses and the softening of the steels. These treatments may be divided into two general classes. The first, called *true* annealing or *full* annealing, is defined according to the A.S.M. "Metals Handbook" as the heating of an iron-base alloy above the critical temperature range, holding it above that range for a proper period of time, followed by slow cooling through the range. The annealing temperature is generally about 100°F. above the upper limit of the critical range, and the time of holding is usually not less than 1 hr. for each inch of section. The objects so treated are ordinarily allowed to cool in the furnace. The other class of treatments, termed *annealing*, consists of heating an iron-base alloy to a temperature below or close to the lower critical range and following this by cooling as desired. This treatment, which is generally termed *process annealing*, is commonly applied in the sheet and wire industry. These two classes of annealing have been distinguished because the results that each produces are quite different, although the treatments accomplish more or less the original purpose of stress relief and softening.

The purpose of these two classes of treatments may be to accomplish one or more of the following:[1]

1. To remove stresses.
2. To induce softness.
3. To alter ductility, toughness, electrical or magnetic characteristics, or other physical properties.
4. To refine the crystal structure.
5. To remove gases.
6. To produce a definite microstructure.
7. To produce grain reorientation.
8. To homogenize the structure.

Annealing Processes.—True or full annealing consists in heating a steel above the upper critical and following this by slow cooling. For a hypoeutectoid steel, which is an aggregate of ferrite and pearlite, full annealing involves, first, heating above the upper critical range to effect complete solution or transformation to austenite. The mass of steel is held at this temperature until complete solution and homogenization is affected and then is slowly and uniformly cooled. The resulting structure, if the heating, soaking, and cooling have been properly carried out, will be lamellar pearlite and free or excess ferrite.

In the ordinary full-annealing treatment, where the cooling through the transformation range is at a very slow rate, the excess ferrite will fully separate from the pearlite. From a practical consideration the production of such a structure requires that the

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[1] A.S.M. "Metals Handbook," p. 3, 1939.

steel be cooled in the furnace and that the rate of cooling be in inverse proportion to the size of the piece. The determination of the exact rate of cooling required to produce complete separation with a given section must be determined by experiment under the given working conditions. Any variation from this rate will alter the way in which the ferrite (and cementite) is precipitated from the austenite. Such variations will produce structures varying from a fine network, where only a small amount of ferrite has precipitated, indicating a relatively rapid rate of cooling, through intermediate stages showing coarser networks and partial separation, to structures showing complete separation of the excess ferrite, which denotes very slow rates of cooling.

Annealing of eutectoid and hypereutectoid steels presents a somewhat different situation owing to the percentage difference in structural composition. In these steels the pearlite content is of major importance since the amount of excess cementite in the usual commercial grades does not exceed approximately 5 per cent. Another factor in the operations is the rapid rise of the A_{cm} critical, which for full annealing would demand an extremely high temperature. Since grain growth increases with an increase in temperature above the critical range it has been found commercially desirable to anneal these steels between the $A_{1.3}$ and the A_{cm}, with the temperature usually close to the $A_{1.3}$. The structure under this treatment will consist of excess cementite in the grain boundaries and pearlite.

Heating for Annealing.—Heating, of course, should be as slow as is practicable and should be uniform. Rapid or nonuniform heating would induce additional strains over those which might already be in existence in the section. Uniformity of heating is particularly important in the case of large objects, where the temperature of the interior tends to lag behind that of the exterior. The approach to the final annealing temperature should be slow enough so that the interior can be brought up to it without carrying the exterior too far beyond, because too high a temperature is injurious and would also tend to recoarsen the grains.

Annealing Temperature.—The full annealing operation has for its basic requirements the production of an entirely new crystalline structure of either the smallest grain size attainable or of some required larger grain size. Concurrently this treatment will also eliminate all internal stresses. As has been previously shown, the formation of this new structure occurs as the steel is heated beyond the upper critical. As the temperature is increased beyond this point of complete refinement the grain size will tend to coarsen and, at a sufficiently high temperature, the coarsening may be excessive. There must, therefore, be a limiting full-annealing temperature consistent with the steels being treated. Practically, this temperature is determined by locating the Ac_3 end temperatures and allowing a given leeway, which in the accompanying chart amounts to 100°F. With the so-called *grain-controlled* steels this recommended range is considerably more flexible.

The range of temperatures in Table 1-IX is recommended by the A.S.T.M. for steels of the carbon ranges shown, with the usual moderate manganese content. For steels with manganese over 0.75 per cent, temperatures slightly lower would suffice.

TABLE 1-IX

Range of Carbon Content	Range of Annealing Temperature
Less than 0.12 per cent.......	875 to 925°C. (1607–1697°F.)
0.12 to 0.29 per cent..........	840 to 870°C. (1544–1598°F.)
0.30 to 0.49 per cent..........	815 to 840°C. (1499–1544°F.)
0.50 to 1.00 per cent..........	790 to 815°C. (1454–1499°F.)

The annealing temperature ranges, as can be seen, vary somewhat with the carbon content; the higher the carbon content, the lower will be the annealing temperature. In the case of large objects it is found advantageous to carry the temperature of the exterior somewhat higher so that the interior may be brought up to an effective annealing temperature. Therefore, for each carbon range there is a definite range of annealing temperatures. The upper temperature limit applies to large objects and also to the lower part of the carbon range given.

The effect of the annealing temperature on the physical properties of the steel is also of considerable importance. The general trend of steels annealed at different temperatures and slowly cooled in the furnace is that as the temperature rises there occurs a softening, starting at a temperature somewhat below Ac_1 and reaching the maximum softness as the Ac_3 is passed. On further increase of temperature above the Ac_3, the tensile strength is not materially changed but the ductility is found to drop very rapidly. This trend is probably due to the coarsening of the structure, which makes it necessary, in order to obtain the best combination of strength and toughness, to limit the temperature of heating above the Ac_3 to avoid the coarsening effect. The impact resistance of the material follows the same trend as the ductility but decreases at a somewhat faster rate. A coarse structure, however, is desirable in the case of rough machining operations, since it affords better machining on this type of cut.

Cooling from the Annealing Temperature.—After the object has been held at the annealing temperature for a length of time that will produce as uniform a temperature as is possible with complete refinement of the grain, relief of internal stress, and sufficient homogenization, it should be cooled in such a way as to suit the carbon content and give the required structure and, hence, the physical properties desired.

To obtain desired softness and ductility in the steel, it should be cooled so that there has been sufficient time for the complete precipitation of the ferrite or cementite and pearlite. Rapid cooling would retain the fine structure desired but would harden the steel and deprive it of its ductility. This would, of course, defeat the purpose of annealing. To accomplish that purpose, the steel should be cooled with the furnace in which it is heated until its temperature has fallen below the critical range.

The rate of cooling allowable will vary somewhat with the carbon content. In general, it may be said that the higher the carbon content, the slower must be the cooling. For instance, a steel containing not over 0.15 per cent carbon may be quenched in water, thereby increasing its tensile strength, and it will remain quite ductile. The reason for this is the relatively wide temperature differential between the upper and lower criticals and the predominance of ferrite in these steels, which will, even with fast rates of cooling, tend to precipitate. On the other hand, with higher carbon steels a rapid cooling rate can be used only at the sacrifice of ductility.

The obvious objection to cooling the steel in the furnace is that it may materially slow up production. However, experience indicates that neither of the extremes regarding softness, ductility, strength, and elasticity is generally required, so that it is sufficient, from the standpoint of both quality and economy, to cool the steel, let us say, to 1000°F., or until it has reached a black heat, after which it can be removed from the furnace and cooled in the air.

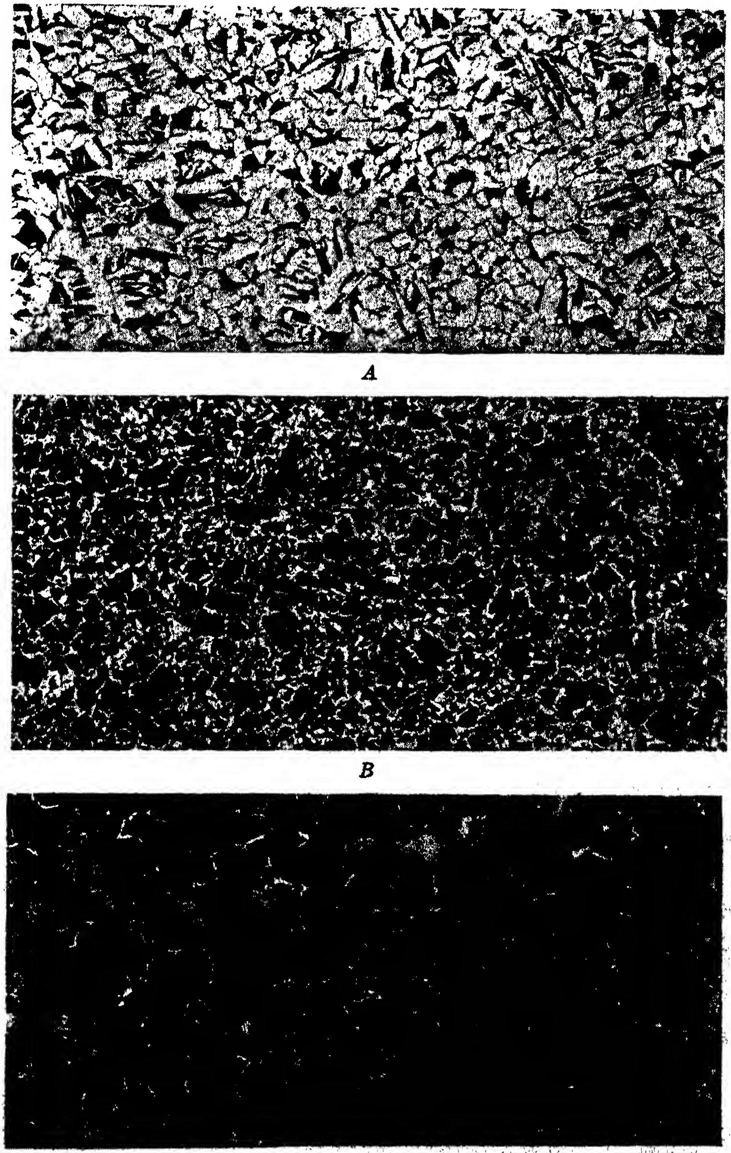
✓NORMALIZING

Normalizing is defined in the A.S.M. "Metals Handbook" as the heating of an iron-base alloy to approximately 100°F. above the critical range, followed by cooling to below that range in still air at ordinary temperature. Commercially, it has for its purpose the production of uniform conditions on materials that have been treated in various ways and also, in the case of castings or forgings, the breaking down of the coarse structure that has developed at the high temperatures used.

The Process of Normalizing.—The process consists of, first, heating above the upper critical to effect complete solution of the original structure with the formation of austenite. Theoretically, it is possible to effect complete solution by heating to just above the upper critical; but in cases where the stress produced by rolling, forging, or casting must be completely eliminated and where the structure produced is really undesirable, *i.e.*, where it is necessary that a new equiaxed grain structure be produced in the metal, it would require an extremely long time at this temperature to effect solution and bring about the desired results. At a higher temperature this solution will take place much faster. With subsequent air cooling from this temperature, the structure✓ is found to be somewhat different from that obtained with furnace cooling.

Rates of Cooling and Their Effects.—For an accurate understanding of the structures obtained, it is necessary to make a more detailed study of the effects of cooling rates on the precipitation of the proeutectoid ferrite or cementite and pearlite.

With the slow rate of cooling (furnace) from above the critical, the hypoeutectoid steel will precipitate ferrite from the austenite between the criticals Ar_3 and Ar_1. The ferrite thus precipitated will form a coalesced network around the remaining austenite. At the Ar_1 critical, the austenite, which now contains 0.85 per cent of carbon, will transform to coarse-lamellar pearlite. With increased rate of cooling the lamellar plates of the pearlite will become progressively finer and more of them will be packed into the same space until, finally, a cooling rate will be reached that



A

B

C

FIG. 2-IX.—*A*, Structure of 1020 carbon steel after normalizing at 1650°F. × 100. Nital etched. *B*, Structure of 1040 carbon steel after normalizing at 1550°F. × 100. Nital etched. *C*, Structure of 1060 carbon steel after normalizing at 1500°F. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

The photomicrographs in Figs. 2-IX and 3-IX, show the structures of normalized hot-rolled steels containing 0.20, 0.40, 0.60, 0.85, and 1.20 per cent of carbon, all of which have been etched in nital and were taken at 100 ×. (Compare with Figs. 1-VII and 2-VII.)

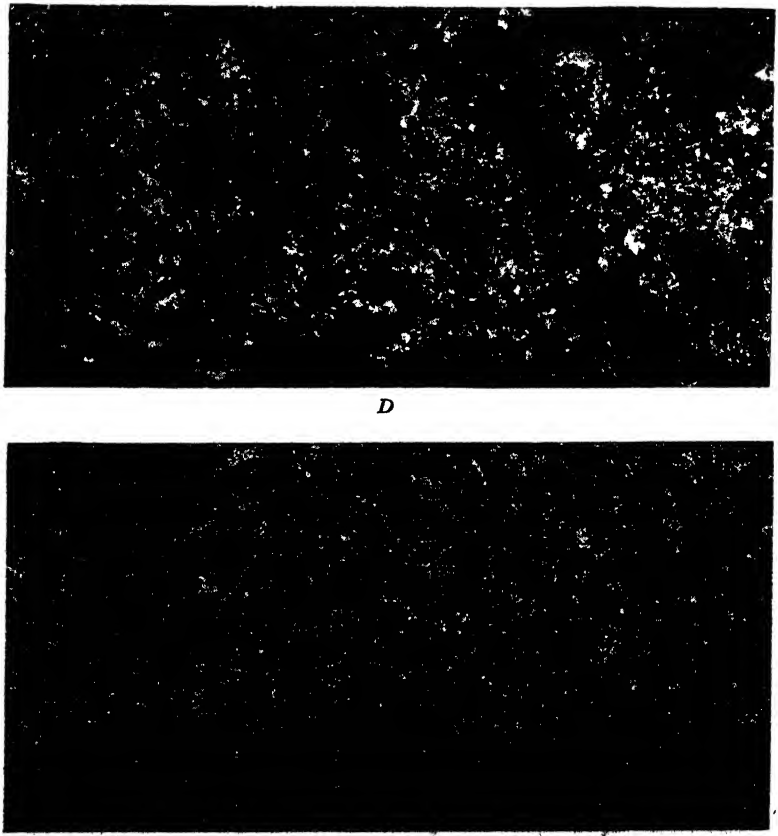


FIG. 3-IX.—*D*, structure of 1085 carbon steel after normalizing at 1450°F. × 100. Nital etched. *E*, structure of 10120 carbon steel after normalizing at 1650°F. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

Normalizing versus Annealing.—Data in Table 2-IX (page 178) show the physical properties of four different carbon steels as they are affected by the two different treatments. The tensile tests were made on standard 0.505-in. tests taken from the axis of 1-in. round bars. The Brinell values were taken from the tensile bars and are, therefore, representative of the internal hardness rather than the surface or skin hardness.

As shown in Table 2-IX, the effects of annealing when compared with those of normalizing are to produce lower tensile and yield strengths and to lower them more in the high carbon than in the low carbon steels. The elongation and reduction of area under the same comparison show a much greater change in the higher carbon steels.

SPHEROIDIZING

Spheroidization is defined, according to the A.S.M. "Metals Handbook," as any process of heating and cooling that produces a rounded or globular form of carbide. In hypoeutectoid and eutectoid steels, this involves the divorcing or the agglomeration of the cementite lamellae of the pearlite from the ferrite. In hypereutectoid steels, it is the change of both the cementite of the pearlite and the proeutectoid cementite into the globular form.

Purpose of Spheroidizing.—Spheroidization is applied to hypoeutectoid carbon steels for the general purpose of meeting given strength requirements, either in relation to annealing operations or as a preliminary to subsequent heat-treating operations. This treatment as applied to carbon steels of from approximately 0.75 per cent up involves not only the improvement of structure as it will affect the properties of the material on subsequent heat-treatment, but—what is of equal or greater importance—the improvement of machinability, as well. Machinability is of particular importance in steels of high carbon content since they are usually so hard that, were it not for spheroidization, they would be practically unmachinable. Spheroidization produces maximum ductility and softness at the expense of the strength and elasticity, but the two latter are but slightly impaired.

A spheroidized steel has a further desirable property—"wearability." The spheroidized structure, which contains a ferrite matrix with small globules of hard cementite—the hardest constituent of steel—is, therefore, excellently suited to resist abrasion. Under abrasive action the soft ferrite wears away, leaving the cementite particles exposed to meet the abrasive object in contact. The ferrite matrix makes an excellent reservoir for lubricants, which are supplied to the cementite bodies to break the metal-to-metal contact.

If two steels of the same carbon content, one in the spheroidized condition and the other in the lamellar, the former condition will be the softer. The reason for this is apparent if we consider the geometric axiom that the figure having the greatest volume for its surface area is the sphere. Hence, the spheroidal carbides present less area to the hardness-penetration ball than do the lamellar carbides of the same cubical content and will, therefore, be the softer of the two structures.

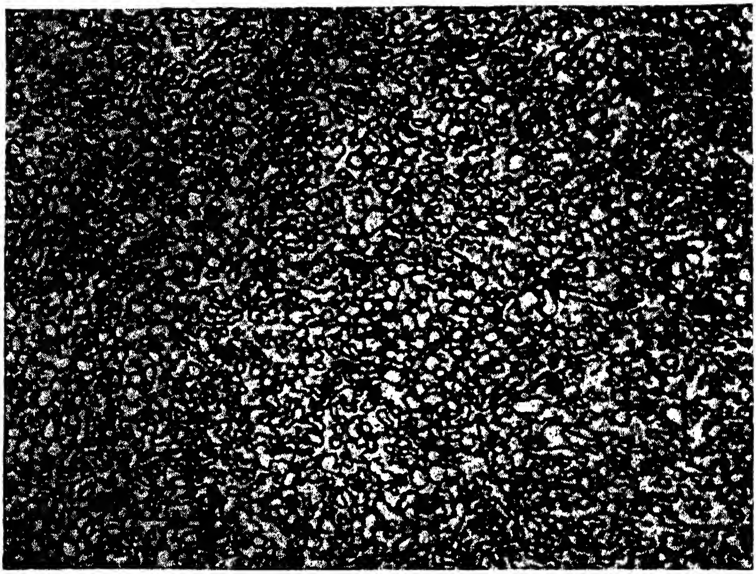


FIG. 4-IX.—Spheroidized cementite. × 1,000. (*Courtesy of Bethlehem Steel Company.*)

Spheroidizing Treatment.—The spheroidizing treatment most widely used consists in prolonged heating of the steel at a temperature in the neighborhood of, but generally slightly below, the critical temperature range, usually followed by relatively slow cooling. Both the cementite in pearlite and the free proeutectoid cementite will spheroidize into a structure such as that shown in Fig. 4-IX.

Spheroidizing may be applied to any steel containing cementite or similar carbides as in alloy steel, whether they be cold-worked annealed, normalized, or quenched and tempered. Some steels are exceptionally difficult to spheroidize but can be spheroidized by increasing the time at the proper temperature or by oscillating the temperature back and forth—in and out—through the lower critical. It is interesting to note that sluggishness or unresponsiveness to the treatment, in general, increases with carbon content. Eutectoid steels, in particular, offer unusual resistance to spheroidizing.

Certain preliminary operations are particularly amenable in accelerating the globular coalescence of the iron carbide. These are as follows: (1) cold work, (2) obtaining as fine a pearlite as possible in normalizing, (3) lead quenching, and (4) hardening. The latter two will be discussed in a later section.

A recommended practice for spheroidizing is to normalize at the temperature recommended for the per cent carbon steel being treated, using the proper length of time with respect to the section size; cool in air to about 800°F.; reheat to between 1300° to 1350°F. for 1 hr. per in. of diameter of thickness; and cool in the furnace to at least 900°F., from which temperature the steel may be reheated for further treatment or cooled to room temperature in air.

MACHINABILITY

The property of machinability belongs to materials that are amenable to machining operations. Good machinability is generally associated with ease of obtaining smooth-finished surfaces and with long tool life at high cutting speeds. The ideal machining condition in steels is obtained with structures that produce low hardness values along with a tendency toward brittleness. This condition can be approached by altering the microstructure; by cold drawing, which decreases the ductility; by coarsening the grain; by increasing the phosphorus or nitrogen content of the ferrite, thereby increasing the brittleness; by introducing nonabrasive inclusions, such as sulphides and selenides, and by alloying additions that increase the rate of work hardening and thus embrittle the chip.

The most recent innovation in improved machinability is the use of lead additions in screw stock.[1] The reason for the behavior

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[1] Nead, Sims, and Harder, Properties of Some Free-machining, Lead-bearing Steels, Part 2, *Metals and Alloys*, **10**, 109–114, (1939).

of lead in this capacity has not been definitely established; however, it seems not to fall into any of the above categories but to be additive to the effects produced by factors that do fall into those categories. It is evident that lead in screw stock increases the machinability, with no noticeable effect on the mechanical properties.

Machinability of steels may be classified[1] (exclusive of high sulfur screw-stock steels) into four groups, depending upon the carbon content. The low carbon steels, of about S.A.E. 1020 grade, machine best after normalizing. At their best, these steels are difficult to machine because the ferrite, which is a continuous, soft, and ductile material, is "sticky" and "gummy" to cut. In the higher carbon group, say 0.30 to 0.40 per cent, the stock should be normalized high enough to coarsen the austenitic grain and then cooled rapidly, to give a fine pearlite with a minimum of free ferrite. In steels containing 0.50 to 0.60 per cent of carbon, an annealed structure with lamellar pearlite is best, making sure that the pearlitic grains are not entirely surrounded with ferrite. Finally, in the eutectoid steels a spheroidizing anneal is required. Microstructure is, according to findings, of prime importance to machinability.

All steels containing a large amount of free ferrite are soft and easily cut, causing little wear on the cutting tool. They are tough and easily deformed, and their chips tear away from the stock, leaving a rough surface. However, the chips tend to adhere to the cutting edge, where, as a result, excessive frictional heat is created, tending to draw the temper of the tool. The conventional means of retaining the good qualities of ferrite and eliminating the bad machinability is to introduce into the stock manganese or selenium sulfide—a solid, nonmetallic inclusion.

The probable effect of such particles,[2] as an aid to both tool life and surface finish, is that each cavity containing such an inclusion behaves as a minute notch, since the reinforcement of the soft inclusion is practically nil. Because of these highly localized stress concentrations just ahead of the cutting edge of the tool, the chip comes away readily and the material behaves—

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[1] Some Factors Affecting the Machinability of Steels and Other Alloys, *Metal Progress*, **29** (No. 1), 31–34 (1936).

[2] "U.S.S. Carilloy Steels," Machinability of Alloy Steels, pp. 183–186, Carnegie-Illinois Steel Corp., Pittsburgh, 1938.

at least, with respect to the cutting tool—as if it were brittle. The spheroidized cementite particles of annealed high carbon steels probably act in the same capacity.

PATENT ANNEALING

Use in the Manufacture of Wire.—This type of annealing is used in the wire industry—either as a finishing treatment, whereby certain desirable physical properties may be obtained, or as a treatment prior to further wire drawing. This latter is particularly important in the case of higher carbon steels where it is necessary for the constituents to be in such a form that they may have sufficient ductility for deformation. In low carbon steels, say under 0.30 per cent of carbon, the constitutents, particularly pearlite, are in so small an amount that they are not noticeably detrimental to the cold-working properties of the metal. The free ferrite, which is predominant in these steels, is plastic and is easily deformed. Areas of free cementite, on the contrary, will resist deformation. The plasticity of pearlite, which consists of alternate bands of ferrite and cementite, will, therefore, depend upon the thickness of these plates and upon their relationship to the axis of the wire. Thus, pearlitic areas parallel to the axis will offer less resistance to drawing than will areas perpendicular to the axis. It is also found that the seriousness of this resistance is dependent upon the thickness of the cementite plates. Thin plates or fine pearlite will not offer great resistance to drawing, regardless of their position with reference to the axis.

The most suitable structure for drawing high carbon steel wire to obtain maximum tensile strength should consist of fine pearlite. However, for maximum workability it is necessary to utilize the ductility of the spheroidized structure. The patenting treatment that develops this fine pearlite consists of heating the iron-base alloy above the critical range, effecting complete solution, and following this by fairly rapid cooling in either an air, salt or lead bath. The rate of cooling should be just rapid enough to produce fine pearlite and, when the steel is slightly hypo-eutectoid in composition, it will have a hardness of approximately 35 on the Rockwell C scale. Upon cold drawing, followed by a "stress-relief" anneal, the tensile strength may be increased to

300,000 p.s.i., with the drawn wire still ductile enough to stand wrapping around its own diameter.

Processing the Rod.—Now that the structural requirements of high carbon spring wire are understood, let us briefly study the operations necessary in drawing the green rod to say 15-gauge spring wire. The usual specifications for spring steel will range from 0.50 to 0.70 carbon, 0.90 to 1.20 manganese, 0.040 max. phosphorus, and 0.045 max. sulfur. The hot-rolled rods, after rolling, are cleaned of iron-oxide scale by pickling in an acid-inhibited bath. They are then rinsed and given a "sull," or wet-rust coat and a lime coat and then are baked. The "sull" coat acts as a carrier of the drawing lubricant and the lime coat assists in neutralizing any residual acid and facilitates lubrication during drawing. The green rod so processed is then drawn to a smaller diameter, prior to patenting, as very few sizes of spring wire can be drawn directly from patented rods. The green rods will have a coarse-grained structure, with large grains of lamellar pearlite, owing to the fact that the finishing temperature is generally rather high (1800° to 1900°F.), and it seems desirable in the light of recent investigation to use a coarse-grained steel. The green rods must, therefore, not be drawn, or "ripped," too drastically to the patenting size because of the deficient ductility when they are in this condition. Overdrawing would result in internal and external ruptures.

The Patenting Operation.—The spring steel rods or wire are now ready for heat-treatment to condition them for subsequent drawing. The patenting operation consists in passing the rods, single strand—usually a number of single strands at the same time—through a suitably designed furnace whose temperature is automatically controlled. The patenting temperature generally ranges around 1800 to 1850°F., although it may be over a wide range above the critical temperature, since this is essentially a time-temperature process. If the temperature is high, the time can be low and, if the temperature is low, the time will be high. A definite schedule of speed of passing through the furnace must be set up for a given temperature and a given wire or rod size. The time at the patenting temperature must be sufficient to allow complete transformation to austenite and also to allow the austenite grains to grow. If the austenite grain of the original steel is coarse, it will tend to coarsen further.

The Rate of Cooling.—The control of the rate of cooling from the patenting temperature must now be considered, since this is the part of the operation that determines the structure. The cooling must be at such a rate that there is obtained, in the ideally patented structure, fine pearlite with a minimum precipitation of free ferrite or cementite in the grain boundaries (depending upon the carbon content being worked). This structure will give the greatest freedom to yield to deformation in any direction and will retain maximum toughness with maximum cold drawing.

Fig. 5-IX.—Air patented structure of five gauge rod. × 500. Ladle analysis 0.64 = 1.10 = 0.020 = 0.035 = 0.12; patenting furnace temperature 1850°F.; speed, 15.7 ft. per min.

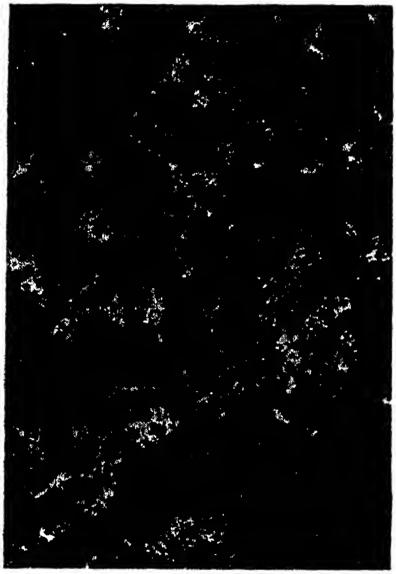

Fig. 6-IX.—Lead patented structure of ¼-in. rod. × 500. Ladle analysis 0.71 = 0.65 = 0.021 = 0.029 = 0.22; patenting furnace temperature 1850°F;. lead temperature, 1115°F.; speed, 11.78 ft. per min.

Figures 5-IX and 6-IX show the typical fine pearlitic structures that are produced by air and lead patenting. The differences in the relative fineness of the lamellae of the pearlite produced by the two rates of cooling are quite evident. In general, this difference in patented structure will reflect itself in the properties, on subsequent cold drawing. The properties of lead-patented, cold-drawn wire are slightly superior to those of air-patented, cold-drawn wire.

It can be noted from the comparative analyses that the manganese content is somewhat higher in the air-patented than in the lead-patented steel. In general, it is found[1] that when air quenching is employed it is desirable to increase the manganese above the usual range so as to increase the hardening characteristics of the steel.

Properties of the Finished Wire.—Following patenting at 1800°F., the finished wire size is obtained by drawing. The total reductions taken in reducing from patented stock to finished wire is determined by the tensile strength, resilience, and ductility required by the spring fabricator. The fabricator usually requires a substantial amount of residual ductility in the purchased wire, so there must be a proper balance between carbon content and total reduction. Eighty per cent total reduction in cross-sectional area from patent stock to finished wire is a conservative limit for a reduction schedule.

Fig. 7-IX.—Microstructure of 15½ gauge spring wire, 0.59 per cent carbon and 1.08 per cent manganese reduced in four drafts from 0.142 patented stock to 0.067 in. Tensile strength, 228,900 p.s.i. Longitudinal section. × 1,000. 2 per cent nital etched. (*Made in the laboratories of Keystone Steel & Wire Company, Peoria, Ill.*)

Figure 7-IX shows the photomicrograph of a longitudinal section of 15½-gauge spring wire of normal quality, medium tensile strength and resilience, and high residual ductility. The

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[1] Fisher, J. P., Cold Drawn Spring Wire, *Wire*, **9**, 287–297, (1934).

microstructure is typically fibrous and is composed chiefly of fine pearlite with a small percentage of coarser lamellar pearlite. The effect of drawing on coarse lamellar pearlite which is perpendicular to the wire axis can be noted by closely examining the small patch of coarse pearlite in the center of the photomicrograph. The high pressure and the hardness developed in the surrounding areas have resulted in a bending of the plates.

ANNEALING FOLLOWING COLD WORK

The second class of treatments that are termed *annealing* is employed in a large group of commercial steel products that have been plastically deformed at such temperatures and at such rates that no recrystallization occurs during the working process. This class of products includes wire and rods, cold-drawn and cold-rolled tubular products, sheet, and many others. The effect of such cold, permanent deformation on the structure of the material varies with the degree of deformation. At low degrees of deformation, lines appear within the grains (slip lines) that denote the yielding in the material. At higher degrees of deformation slip lines appear in more and more of the grains and the grains become elongated in the direction in which the work is applied. The result of such cold deformation is to set up internal stresses that may or may not be harmful, depending upon the requirements; an increase in such strength properties as hardness, yield strength, tensile strength; and a decrease in the ductility.

When it is desired to increase the ductility of such cold-worked metal so that it may be subjected to additional work or so that it will have sufficient ductility and softness for commercial application, it must be subjected to heat-treatment.

From our previous brief discussion of drastic cold work, it was seen that the grains of ferrite and pearlite under the effects of such mechanical work have been elongated and distorted and that the material is under a condition of internal stress. This condition is, of course, extremely unstable and, being unstable, will attempt a return to the more stable state at the first opportunity. Heating to temperatures above the upper critical and slowly cooling would, of course, bring about recrystallization of the grains and relieve the stressed condition (pearlite is considered the commercially stable phase, while actually the spheroidized condition is the closest approach) but recrystallization may also be brought about by heating to below or close to the lower limit of the critical range followed by cooling as is desired. This treatment is commonly called *process annealing* and is generally used at temperatures ranging from 1000°F to 1300°F. (540° to 705°C.).

The Effect of Annealing on Cold-worked Steels.—Process annealing of cold-worked steels brings about a change in the properties that ultimately approaches the characteristics of the material prior to working.

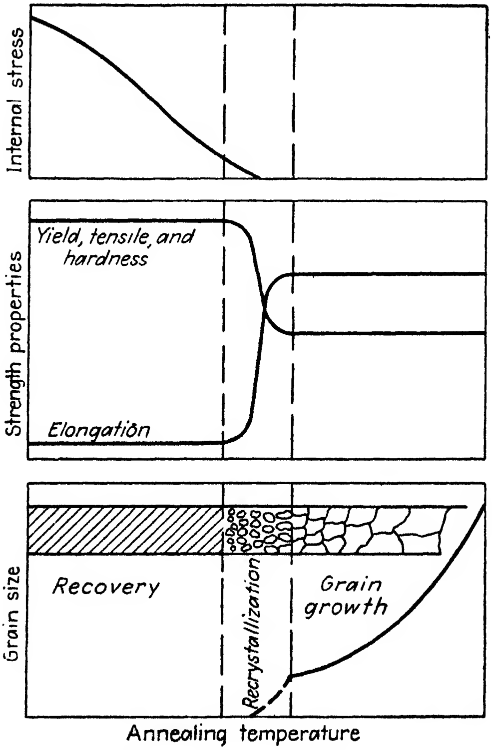

FIG. 8-IX.—Schematic representation of recovery, recrystallization, and grain growth.

The processes that occur on such treatment are of three distinguishable types: (1) recovery, (2) recrystallization, and (3) grain growth, as illustrated in Fig. 8-IX. Recovery is a change in the properties, particularly those determined by the state of internal strain, of the cold-worked steel at low annealing temperatures that is unaccompanied by any detectable change in the microstructure of the cold-worked metal.

Recrystallization is the development of an entirely new grain structure by heating the cold-worked steel to a sufficiently high temperature or by heating it for a sufficiently long time at a lower temperature. The general trend of property change brought about by recrystallization can be seen in Fig. 8-IX which shows that the strength and hardness decrease with a rapid increase in elongation.

The recrystallization process is a very important practical phenomenon, since by this means it is possible completely to remove the effects of cold work and control the grain size and, thereby, the behavior of the material on further working. The factors governing the grain size after recrystallization are as follows:[1] (1) the amount of cold deformation, (2) the temperature to which the cold-worked metal is heated, (3) the time during which the metal is held at this temperature, and (4) the original grain size.

Grain growth of steel after recrystallization is nothing more than continued growth of the recrystallized grain.

It should be emphasized that the three processes are not instantaneous but are dependent upon a time factor. Thus, when the temperature used is high, the processes proceed with such speed that they appear instantaneous, while at low temperatures the time factor is very evident and makes the processes more distinct.

The time factor as correlated with the degree of deformation and recrystallization temperature under simple cold-working conditions, according to Mehl, is as follows: For constant annealing temperatures, shorter times are required for recrystallization at higher degrees of deformation; while with increasing temperatures, lower degrees of deformation and shorter times are required for recrystallization.

The effect of original grain size becomes evident when cold-working steels of two different grain sizes. It is found that the amount of deformation necessary to produce a given amount of strain increases as the grain size increases. In order to obtain similar recrystallization behavior at a given temperature for two

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[1] MEHL, R. F., Recrystallization, A.S.M. "Metals Handbook," p. 207, Cleveland, Ohio, 1939.

FIG. 9-IX.—Grain growth in tapered tensile specimen due to recrystallization after cold working. (*From A. Sauveur, "Metallography and Heat-treatment of Iron and Steel."*)

steels, one large- and one small-grained, it is necessary to effect large deformation in the large-grained specimen and small amounts of deformation in the small-grained specimen.

Another important factor in the recrystallization of cold-worked steels is the factor of critical grain growth. During the drawing operations of certain products it is found that intermediate anneals are necessary so that the products will be sufficiently soft for further reduction to the desired specifications. It has been found that if the drawn product has received a reduction of between 7 and 15 per cent and the temperature of the annealing furnace is in the usual annealing temperature range of low carbon steels, an abnormal grain growth will result. This effect, originally observed by Sauveur, is illustrated in Fig. 9-IX, which shows the microstructure through the cross section of a tapered tensile specimen that has been annealed after cold working. It will be noticed that the coarsest grains do not occur where the metal has been worked the most severely, but do occur at a

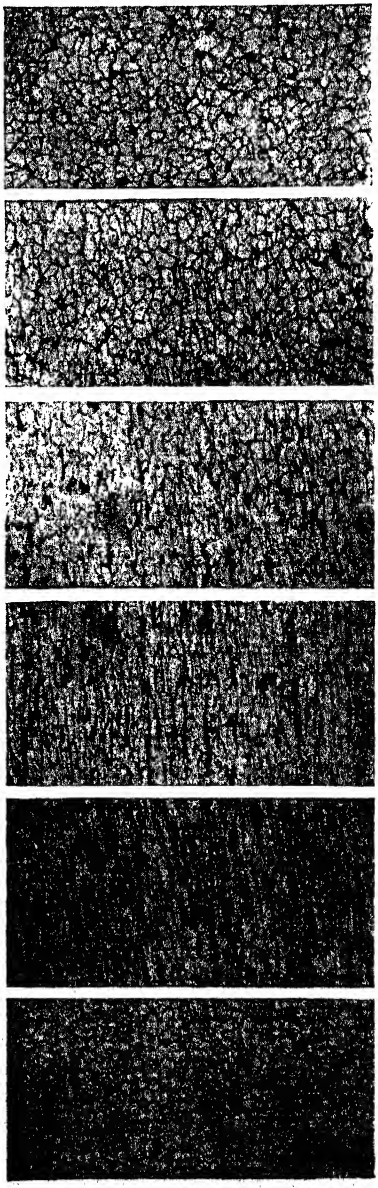

Fig. 10-IX.—Structural evolution of 12½ gauge wire from five gauge green rod. 0.08 per cent carbon 0.32 per cent manganese. × 120. *(Photomicrographs made in the laboratories of Keystone Steel & Wire Company, Peoria, Ill.)*

point where the strain is between the criticals limit. In practice, the products are reduced 20 per cent or more before annealing, thus avoiding the possibility of critical grain growth and producing a satisfactory structure. This same effect may occur in all types of cold-drawing operations.

The series of photomicrographs in Figs. 10-IX and 11-IX shows the effect of cold work, and of annealing and normalizing following cold work, on the structure of low carbon steels.

Figure 10-IX presents the successive structural changes that take place when an 0.08 per cent carbon steel is cold-drawn from a 0.218 in. diameter hot-rolled or green rod in four drafts to a 0.0965 in. diameter wire. Note the difference in flow of the ductile ferrite grains (white) and the harder, more resistant pearlite grains (black), following each successive draft. The resultant effect of cold work and a 1-hr. anneal at 1320°F. on the ultimate tensile strength is also shown.

Figure 11-IX shows the effect of temperature, with constant heating time, on the recrystallization of a low carbon killed steel that has been cold-reduced 75 per cent. Figure 11-VIII*C* shows the microstructure of the cold-reduced metal, the ferrite (white) and the pearlite (black) in the distorted condition. Figure 11-IX*A* shows the same structure after annealing at 1050°F., recrystallization having started. It should be noticed here that the ferrite has started to recrystallize but the distorted pearlite is still in the elongated form. This is as should be expected, for no recrystallization of the pearlite could take place until the temperature had exceeded the critical range. Figure 11-IX*B*, after annealing at 1150°F., and Fig. 11-IX*C*, after annealing at 1250°F., show practically complete recrystallization and growth of the ferrite grains. Figure 11-IX*D*, after com-

TABLE 3-IX

Physical properties affected	Annealing temperature, °F.					
	0	570	840	1020	1110	1200
Yield point, p.s.i.	89,300	86,500	78,600	67,200	40,600	39,400
Tensile strength, p.s.i.	93,000	88,800	84,700	72,700	51,400	51,200
Elongation in 2 in., per cent	6	6.5	13	18	40	40
Reduction of area, per cent	44	55	57	66	73	75

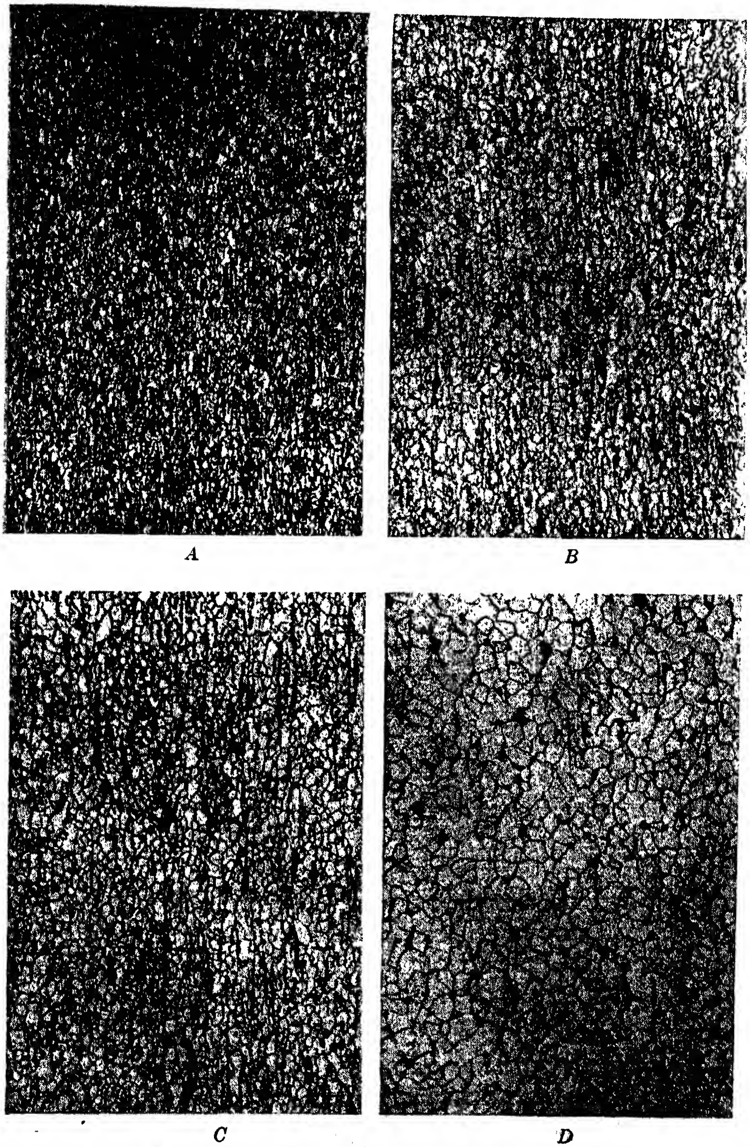

Fig. 11-IX.—*A*, 0.06 per cent carbon steel. Cold-reduced 75 per cent. Annealed 1050°F. × 100. Nital etched. *B*, 0.06 per cent carbon steel. Cold-reduced 75 per cent. Annealed 1150°F. × 100. Nital etched. *C*, 0.06 per cent carbon steel. Cold-reduced 75 per cent. Annealed 1250°F. × 100. Nital etched. *D*, 0.06 per cent carbon steel. Cold-reduced 75 per cent. Normalized 1750°F. × 100. Nital etched.

mercial normalizing at 1750°F., shows complete precipitation of both the ferrite and the pearlite.

The effect of annealing temperature with a constant annealing time of 45 min. on the physical properties of 0.08 carbon wire cold-worked 53 per cent is shown in Table 3-IX.

STRESS-RELIEF ANNEALING

If a steel is subjected to continued cold reduction, the physical properties will continue to change in the manner previously described until a point is reached beyond which there will be no increase. Cold reduction carried beyond this point will result in rupture of the metal. Thus, in cold drawing, the amount of cold working must be kept below this point. This fact makes necessary an anneal above the recrystallization temperature and below the lower critical temperature (Ac_1) when very large amounts of reduction are to be made, as in wire drawing. This is not the case in cold-finished steel, involving rods and bars, where the per cent reduction is low as compared to wire.

TABLE 4-IX.—EFFECT OF VARYING DRAFT ON PHYSICAL PROPERTIES OF S.A.E. x-1335 STEEL[1]

Type of steel	Tensile strength p.s.i.	Yield strength p.s.i.	Elongation in 2 in., per cent	Reduction in area, per cent	Charpy impact, ft.-lb.
Hot-rolled	92,000	57,000	27.5	58.9	
Cold-drawn $\frac{1}{32}$-in. draft	97,000	60,500	23.0	56.6	28.5–28.5
Cold-drawn, $\frac{1}{16}$-in. draft	105,500	68,000	15.5	51.8	27–24
Cold-drawn, $\frac{3}{32}$-in. draft	109,250	72,000	14.5	51.5	26–24
Cold-drawn, $\frac{1}{8}$-in. draft	118,000	81,000	13.0	48.4	21–19

[1] ROBBIN, F. J., Strain Annealing, *Iron Age*, Dec. 14, 1939, pp. 33–38; Dec. 21, 1939, pp. 37–39.

The so-called *stress-relief anneal* involves a low-temperature treatment on materials of the latter class, to produce an increase in tensile and yield strengths, while maintaining greater ductility than is possible in the cold-drawn material. It is possible with this treatment to bring about a wide range of physical characteristics by the use of a combination of cold working in different amounts and low-temperature annealing treatments. Steel in this cold-worked, strained condition can be subjected to a wide temperature range varying up to the point where all the effects of cold working have been removed. This peak temperature is approximately 1250°F. Below this temperature, varying amounts of strain will be released and will produce varying effects on the mechanical properties.

TABLE 5-IX.—IMPROVEMENT IN PHYSICAL PROPERTIES UPON RELEASE OF STRESS IN S.A.E. x-1335 STEEL
Cold-drawn, stress-annealed at 600°F.

Type of steel	Tensile strength, p.s.i.	Yield strength, p.s.i.	Elongation in 2 in., per cent	Reduction in area, per cent	Charpy impact, ft.-lb.
Cold-drawn, 1/32-in. draft	103,500	82,000	20.0	54.2	26–28
Cold-drawn, 1/16-in. draft	115,750	97,000	15.5	50.7	21–23
Cold-drawn, 3/32-in. draft	119,250	98,000	16.0	50.3	20–21.5
Cold-drawn, 1/8-in. draft	128,500	109,000	13.0	45.2	17–19

Tables 4-IX and 5-IX illustrate (1) the effect of draft and (2) the improvement after a stress anneal on the physical properties of a coarse grain (McQuaid-Ehn) S.A.E. x-1335 steel. The annealing temperature of 600°F. represents a condition where considerable stress still remains in the steel and where the tensile and yield strengths are high and ductility low.

Suggested Questions for Study and Class Discussion

1. What are the purposes of annealing steel? Of normalizing? Of spheroidizing? Of patenting?

2. Describe the heat-treating cycle that might be used and the microscopic changes that are brought about in the following:

a. Dead soft-annealed wire having an analysis of 0.21 per cent carbon, 0.44 per cent manganese, low phosphorus, and low sulfur. This wire has been cold-drawn from the green rod to 12½-gauge wire (0.0764 in.).

b. Continuous pot annealing of 30-gauge (0.01003 in.) hot-rolled tin plate, having the following composition: 0.10 per cent carbon, 0.40 per cent manganese, 0.15 per cent phosphorus, and 0.030 per cent sulfur.

c. Continuous normalizing of 31-gauge (0.008928 in.) cold-rolled strip in coil form, having the following composition: 0.05 per cent carbon, 0.30 per cent manganese, 0.015 per cent phosphorus, and 0.035 per cent sulfur.

d. Process or box annealing of the above strip in (*c*).

e. Cold-drawn and annealed tubing, having a nominal O.D of 1¼ in. and a wall of 0.49 in., and an analysis of 0.17 per cent carbon, 0.47 per cent manganese, 0.012 per cent phosphorus, and 0.030 per cent sulfur. This tubing received intermediate anneals, during the drawing, of around 1300 to 1325°F., for maximum softness and ductility and following the final draw was given a low-temperature anneal of 930 to 1000°F. Discuss the effect on the microstructure following each treatment.

f. Annealed cast-steel turbine runner, weighing 63,000 lb., with the following analysis: 0.35 per cent carbon, 0.75 per cent manganese, 0.030 per cent phosphorus, 0.030 per cent sulfur, and 0.35 per cent silicon.

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CHAPTER X

THE HEAT-TREATMENT OF PLAIN CARBON STEELS

REACTIONS OF THE STEEL

Under normal conditions of heating carbon steels, it has been found that the ferrite and the iron carbide of which they are composed react together to form a single solid solution of carbon in gamma iron or austenite. The first reaction on heating is the transformation of the ferrite and carbide of the steel, which are present in a definite proportion (the eutectoid composition) at a definite temperature (the eutectoid temperature), into austenite. On further heating, the excess constituent of the original metal, until it is exhausted, is dissolved in the austenite formed at the eutectoid temperature, the excess constituent being either ferrite, in hypoeutectoid steels, or cementite, in hypereutectoid steels. The process for heating is reversed on cooling. This all takes place under equilibrium conditions wherein we are able to specify the temperatures involved and the proportion and composition of the phases so developed. Equilibrium, therefore, tells us nothing of the type or mode of distribution of these constituents under conditions such as undercooling, a condition of essential importance in heat-treatment.

Theoretically, at least, on slow cooling the steels should transform at the equilibrium critical temperature A_E, but, owing to a definite sluggishness of the metals in reaching equilibrium, it is found that, on heating, the critical point appears somewhat above the A_E and, on cooling, somewhat below it. This lag or temperature displacement, caused by inadequate time for equilibrium to be established, has been denoted by the terms *Ac* and *Ar*. Since there is a displacement of the critical temperature in slowly cooled annealed steels, it is logical to assume that, on increasing the rate of cooling from above the critical range, the critical transformation point *Ar* should be gradually lowered. On this assumption, this question immediately follows: Will the austenite still transform to pearlite as in slow cooling or will there be a different type or mode of distribution of the usual constituents?

To find an answer to that question, let us study, by means of cooling curves, the progress of the transformation of the austenite of a series of small specimens of a 0.50 per cent carbon steel with varying rates of cooling. Figure 1-X shows a series of inverse-rate cooling curves that illustrates the effect of rate of cooling on the temperature of occurrence of the transformation of the austenite. The rate of cooling, which might be denoted by the number of degrees of temperature the specimen was cooled per second, ranges from very slow, as in annealed steels, curve *A*, to very rapid, as in curve *F*.

In regard to the original heating temperature and condition of the steel at that temperature, in the following discussion, unless otherwise noted, the temperatures will be above the A_3 critical for hypoeutectoid steels, above the $A_{1.3}$ point of eutectoid steels and above the $A_{1.3}$ point for hypereutectoid steels, which necessarily means that the steels—with the exception of the hypereutectoid steels—will be completely austenitic. The reason for this exception will be considered later.

The Pearlitic or Eutectic Type of Reaction.—If a plain carbon hypoeutectoid steel is cooled very slowly from the heating temperature, it will, upon reaching the upper critical Ar_3, reject free ferrite from the austenite. Such a transformation involves an internal structural change, the allotropic transformation, in which iron in the face-centered cubic condition changes to the body-centered cubic condition. The amount of free ferrite rejected during the cooling is dependent upon what per cent carbon steel is used. When a temperature of 1337°F. (725°C.), the eutectoid temperature, is reached there is complete transformation of the remaining austenite to lamellar ferrite and carbide or pearlite. Therefore, the decomposition of austenite involves (1) the allotropic transformation of iron and (2) the rejection of carbide at the eutectoid. Any steel that transforms in a like manner may be said to be of *eutectoid* or of *pearlitic* type.

Curve *A* of Fig. 1-X graphically represents the conditions of very slow cooling. At Ar_3, which under these conditions will closely approach A_E, free ferrite is rejected, the amount rejected being roughly proportional to the amount of heat evolved. When the Ar_1 is reached, pearlite will be formed, which for a matter of convenience has been called *coarse* pearlite. At room temperature the microstructure will show roughly 59 per cent pearlite and 42 per cent ferrite.

Cooling curves B, C, D, E, and F, respectively, represent conditions of increased rate of cooling or, in other words, respectively represent depressed temperatures of actual transformations as caused by increasing the rate of cooling. The austenite of which the specimen is composed is retained until some particular subcritical temperature has been reached, at which point the actual transformation takes place.

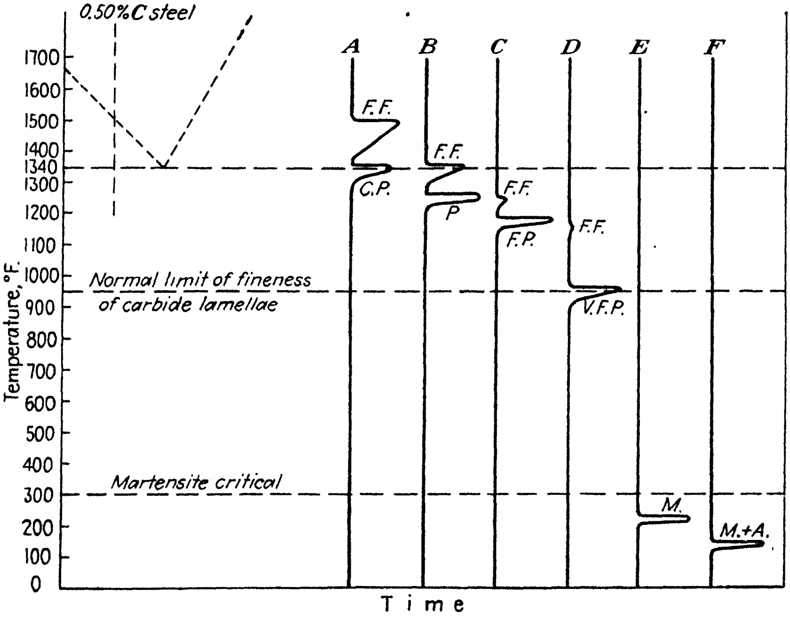

FIG. 1-X.—Graphical representation showing the effect of rate of cooling on the decomposition of austenite. Curve A, slow cooling; curves B, C, and D, intermediate rates of cooling; and curves E and F, rapid cooling (quenched). FF, free ferrite. P, pearlite, the degree of fineness of the lamellae ranging from coarse C to very fine ($V.F.$). M, martensite. A, austenite.

Before discussing each curve, let us draw some general conclusions by observation. Upon increasing the rate of cooling or, in other words, depressing the temperature of actual transformation, we note that the amount of free ferrite that is rejected becomes decreasingly less, with the pearlite becoming more and more predominant until we reach a range where the specimen consists entirely of very fine pearlite. Further, it can be noted that the degree of fineness of the lamellae of the pearlite is increasing until we reach a transformation temperature of between 950° to 1000°F., where it apparently reaches its maximum fineness (curve *D*). In curves *E* and *F*, which represent the extreme conditions of undercooling, a new transition constituent, martensite, makes its appearance and may appear, as noted in curve *F*, in combination with austenite.

It is found in regard to curves *B*, *C*, and *D*, that, even when the cooling has been relatively rapid, there has still been sufficient time for the rejection of some free ferrite—although in a negligible amount—in curve *D*. The remainder of the structure consists of pearlite of varying lamellar spacing. The transformation then, down to about 1000°F., is of the eutectoid or pearlitic type.

It is essential at this point to keep in mind that the temperature is still sufficiently high to allow the carbon to diffuse from the solid solution (austenite) with direct transformation to ferrite and carbide.

Photomicrographs, Figs. 2-X (1) and (2), 3-X (3) and (4), and 4-X (5) and (6), illustrate the effect of subcritical transformation upon the pearlite lamellae and the repective Rockwell C hardness numbers of a 0.78 per cent carbon steel.

The Acicular or Solid-solution Type of Reaction.—If now, as in curve *E*, the rate of cooling is still further increased—an effect that would be produced by quenching rapidly in a cold bath—the temperature will drop past the 1000°F. region so rapidly that the transformation will be prevented from starting until the steel reaches the approximate temperature of the quenching bath. The transformation product at this low temperature, no longer pearlite, is *martensite*, the characteristic constituent of hardened steels. The rate of cooling necessary to just produce the martensitic condition is known as the *critical rate of cooling*. This rate is dependent upon the carbon content, the content of alloying elements other than carbon, and the austenitic grain size.

The formation of martensite is apparently the result of a somewhat different reaction than the pearlitic or eutectic reaction that took place at higher transformation temperatures.

Under the conditions of rapid cooling, austenite will not have an opportunity to transform at the high transformation tempera-

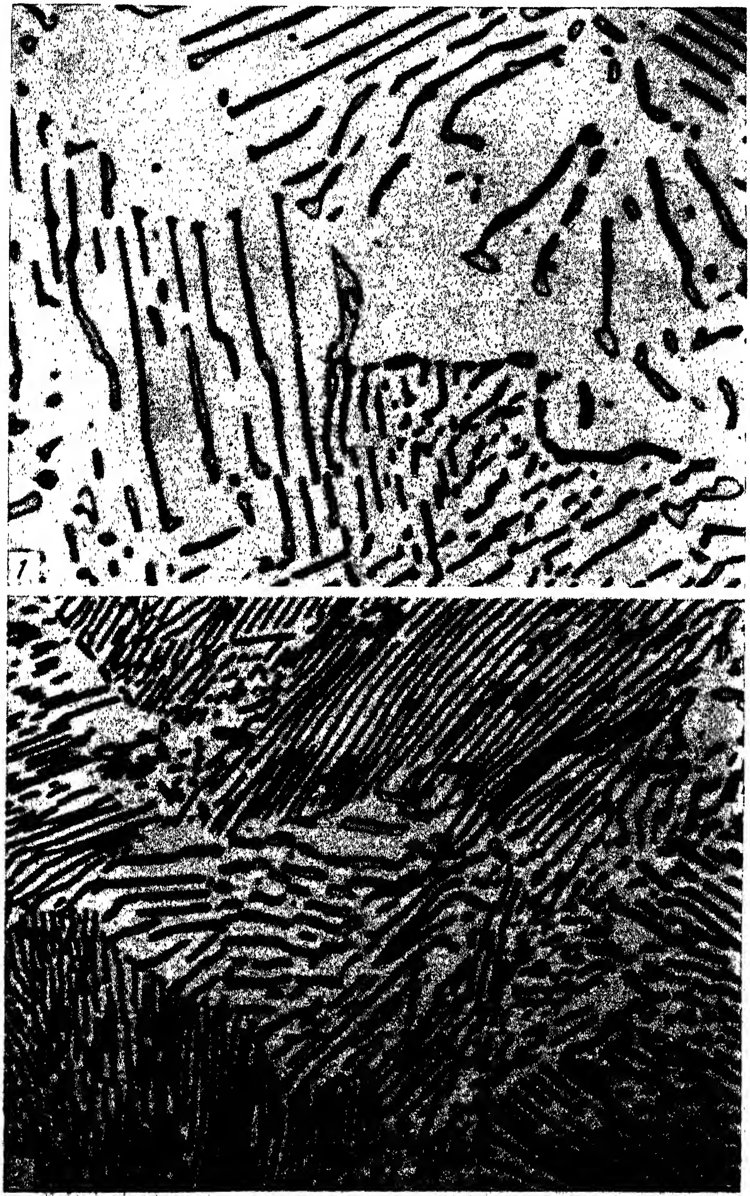

FIG. 2-X.—1, pearlite formed in a 0.78 per cent carbon steel at 1325°F. (720°C.) hardness 7 Rockwell C. × 3,000. 2, pearlite formed in a 0.78 per cent carbon steel at 1300°F. (705°C.) hardness 15 Rockwell C. × 3,000. *(Courtesy of*

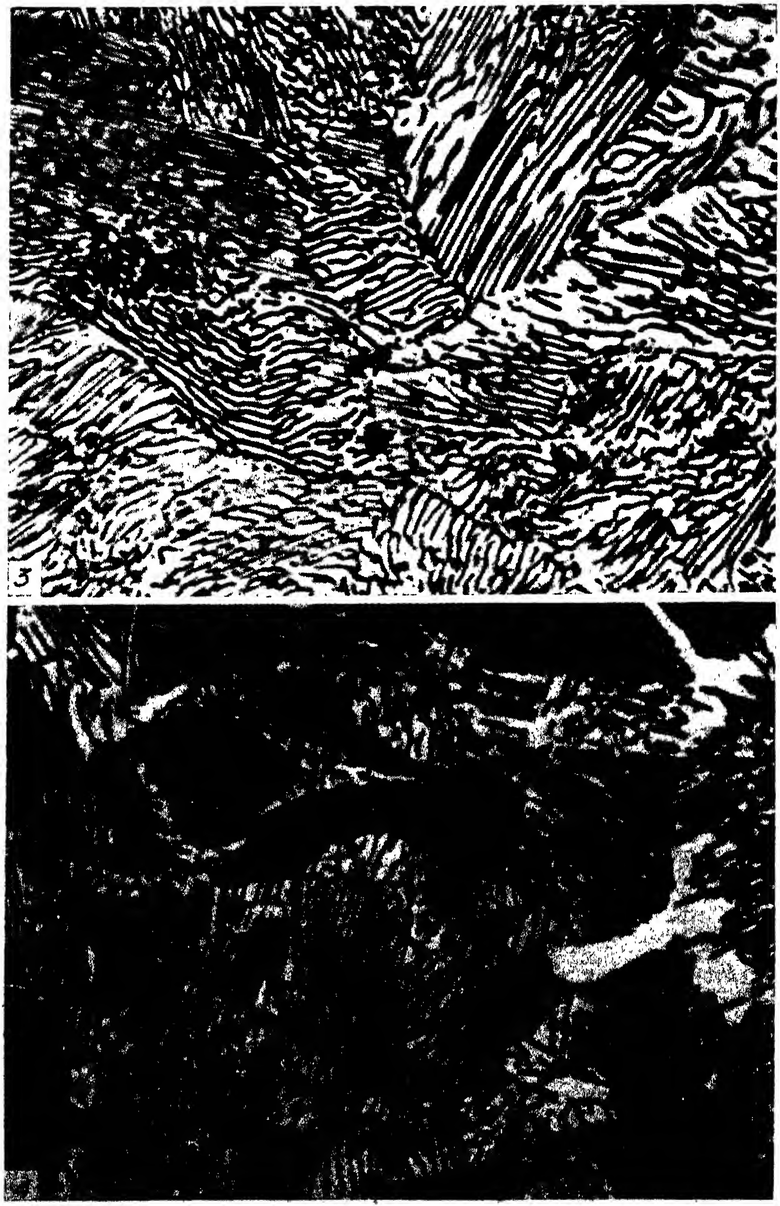

FIG. 3-X.—3, pearlite formed in the 0.78 per cent carbon steel at 1250°F. (675°C.), hardness 22 Rockwell C. × 3,000. 4, pearlite formed in the 0.78 per cent carbon steel at 1200°F. (650°C.), hardness 28 Rockwell C. × 3,000.

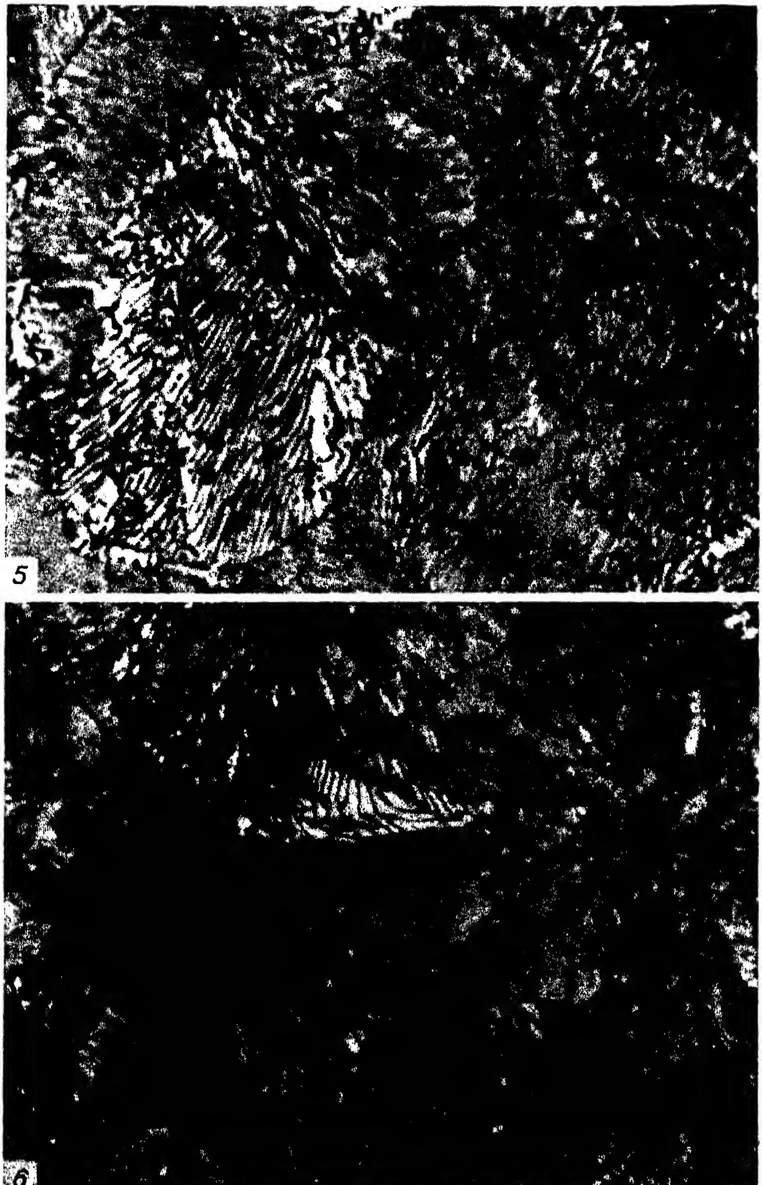

FIG. 4-X.—5, fine pearlite formed in a 0.78 per cent carbon steel at 1100°F. (595°C.), hardness 33 Rockwell C. × 3,000. 6, fine pearlite formed in a 0.78 per cent carbon steel at 1000°F. (540°C.), hardness 36 Rockwell C. × 3,000.

ture to pearlite but will be retarded and will transform at some low temperature to hard martensite. Even at these low temperatures, however, the first step in the austenite decomposition is the allotropic change, the face-centered austenite transforming to body-centered ferrite. The temperature of transformation is so low, however, that the carbon has no opportunity to diffuse; hence, the ferrite produced will be a highly supersaturated solid solution. The degree of supersaturation is evident when we consider that ferrite can retain only about 0.006 per cent carbon in solid solution at room temperature.

MARTENSITE

The Nature of Martensite.—The results of the study of the X-ray diffraction spectrum of martensite have given fairly definite evidence of its nature. Martensite is found to be a body-centered solid solution, which, in the as-quenched condition, is tetragonal rather than cubic. In regard to the tetragonal structure, investigators have found that freshly quenched steels instead of having a perfect body-centered cubic lattice have a tetragonal body-centered lattice. The tetragonal form may best be visualized by considering that the simple body-centered cubic lattice of alpha iron has been somewhat increased in height and the base edges have contracted. The ratio of the height to the base is found to increase from 1.00:1.06 as the carbon content increases from 0 to 1.60 per cent. The distension of the crystal lattice with varying carbon is due to the carbon or carbide held in solid solution. There is no certainty whether the solute is carbon or carbide; however, it is probable[1] that there is a momentary existence of a supersaturated solution of carbon, which almost immediately starts to precipitate carbide particles at a rate dependent upon the temperature.

Most of the investigators believe that the tetragonal structure is an intermediate step in the transformation of austenite to martensite, this form being capable of retaining an appreciable amount of carbon in solid solution. The succeeding transformation takes place either very slowly at approximately room temperature or more rapidly when the steel is tempered at

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[1] Wilella, Guellish, and Bain, On Naming the Aggregate Constituents in Steels, *Trans. A.S.M.*, **24**, 225–226 (1936). Shapiro, C. L., Color Carbon and Aging, *Trans. A.S.M.*, **27**, 666–696 (1939).

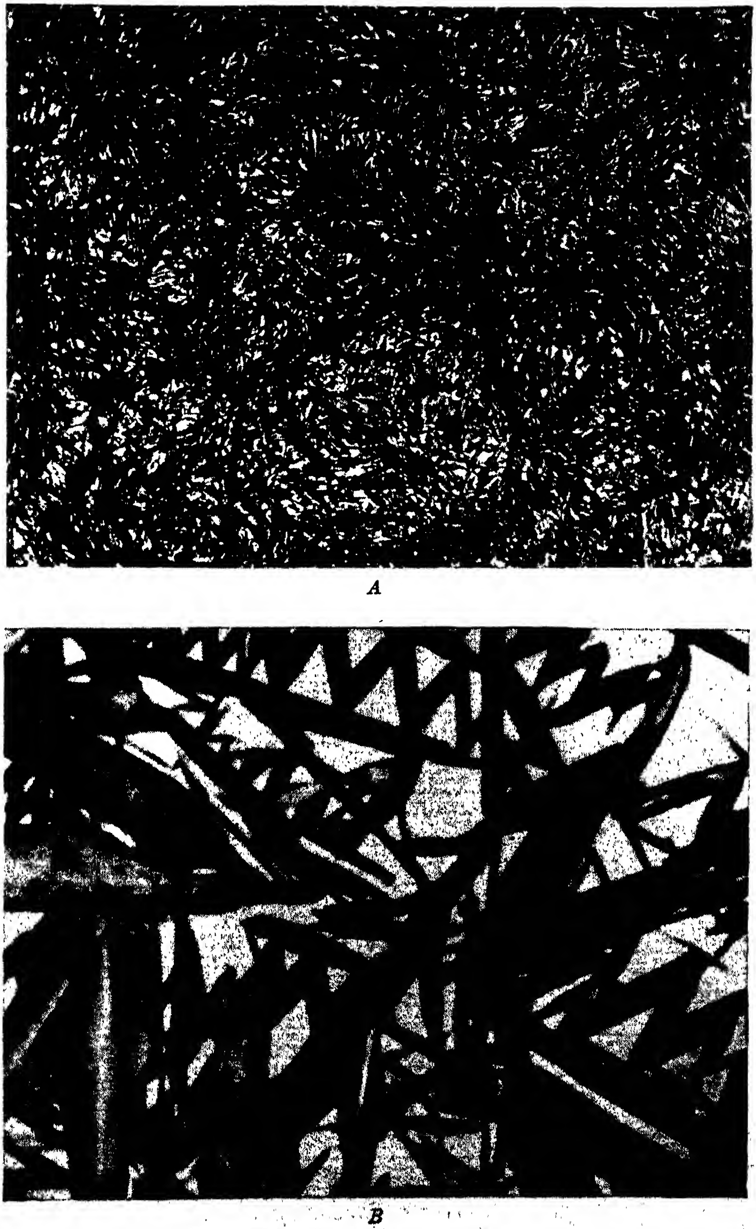

FIG. 5-X.—*A*, typical martensitic structure. × 500. *B*, typical martensitic structure. × 3,000.

210°F. (100°C.). This transformation results in a change of lattice structure from tetragonal to a *temper martensite*, which has a body-centered cubic lattice and is accompanied by a slight increase in hardness. Since this form has very little solubility for carbon at these low temperatures, submicroscopic precipitation will occur.

The Microstructure of Martensite.—Figure 13-X (15) is intended to show the structure of what is termed *white* martensite, in which, probably, the carbide particle precipitation has not occurred to any extent, as is evidenced by the fact that the material does not darken rapidly during ordinary etching for microscopic examination. Figure 5-X, on the other hand, probably illustrates the condition where there has been some precipitation, the structure being termed *tempered* martensite.

The above-mentioned figures, outside their technical differences, show the acicular, or needlelike, pattern with no visible grain boundaries, that always characterizes martensite. The photomicrographs show also that the martensitic needles have formed along certain planes of the parent austenite. Mehl, Barrett and Smith[1] have shown that these needles in both hypo- and hypereutectoid steels form along the same planes as does the ferrite in slowly cooled hypoeutectoid steel. The cementite in slowly cooled hypereutectoid steels forms on an entirely different plane. It is interesting to note, therefore, not only that the martensite of all carbon steels is similar in crystal structure but also that the crystallographic mechanism of its formation is essentially identical with the austenite-to-ferrite transformation in slowly cooled hypoeutectoid steels.

Properties of Martensite.—Of all the transition constituents that will be discussed under the operations of quenching and "tempering," martensite is the hardest and the most brittle, having extremely high tensile strength with little ductility.

The hardness of martensitic steels[2] is determined by its carbon content, regardless of the alloy content in the common S.A.E. steels and regardless of austenitic grain size. The hard-

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[1] MEHL, BARRETT, and SMITH, Studies upon the Widmanstätten Structure. IV, The Iron-carbon Alloys, *Trans. A.I.M.M.E.*, **105**, 215–258 (1933).

[2] BURNS, MOORE, and ARCHER, Quantitative Hardenability, *Trans. A.S.M.*, **26**, 1–36 (1938).

ness values increase rapidly in the lower carbon range, ranging from 48 Rockwell C in a 0.20 per cent carbon steel to 63–65 Rockwell C at 0.50 carbon. With higher carbon contents the maximum hardness values do not vary materially from approximately 66 Rockwell C without a high percentage of alloy content. The retention of austenite in the hardened high carbon steels does not materially affect the hardness values, even though austenite is generally considered to be much softer than martensite.

Table 1-X shows the effect of increasing carbon content on the tensile strength of plain carbon steels as quenched in water and oil, when compared with the annealed and normalized condition. The tensile properties shown are average values obtained on 0.505 test bars machined from ¾ to 1 in. diameter, as treated bar stock.

TABLE 1-X

Carbon content, per cent	Annealed tensile, p.s.i.	Normalized tensile, p.s.i.	As water-quenched tensile, p.s.i.	As oil-quenched tensile, p.s.i.
0.10	48,000	53,000	82,000	72,000
0.20	65,000	70,000	150,000	85,000
0.30	74,000	80,000	238,000	102,000
0.40	88,000	95,000	266,000	119,000
0.60	95,000	108,000	287,000	155,000
0.80	102,000	120,000	311,000	189,000
0.95	106,000	148,000	324,000	220,000

From our previous considerations it should be seen that for maximum hardening of steel the formation of relatively soft pearlite should be prevented. From the above strength and hardness figures it is seen that lower carbon steels do not develop so high a strength or hardness as the higher carbon steels. This is due to the fact that in the lower carbon steels the tendency for the ferrite to precipitate, even with rapid cooling, is rather high. The rejection of even a small amount of proeutectoid ferrite will tend to retard the cooling rate, because of the heat evolved in its formation, and to reduce the strength and hardness.

Another factor may be noted by comparing the strength values of the carbon steels as water quenched and as oil quenched. This involves the fact that, because of the lower cooling rate induced by the oil, with oil quenching plain carbon steels do not produce full martensitic hardness even in the highest carbon contents. With very thin sections and high carbon content it is possible to produce full hardness; however, as the size increases, only partial hardening may be expected. The usual procedure when it is essential to change from water to oil—owing, for example, to the necessity of avoiding distortion—and still obtain full martensitic hardness, is to change the steel composition to a moderately or a highly alloyed steel.

With the hypereutectoid carbon steels, it is usually recommended that, in order to obtain full hardness, the steel be heated to just above the $A_{1.3}$ critical, followed by water quenching. Heating to this temperature does not take all the cementite into solution; but if proper pretreatment, such as normalizing, has been given, the cementite will be well distributed and will tend, if anything, to add to the hardness of the quenched steel. If a temperature exceeding the A_{cm} is used, the structure will be fully martensitic but the steel may be less hard than that quenched at the lower temperature. In addition, the structure will be coarser and the danger of cracking will be greater, owing particularly to the reprecipitation of brittle cementite.

Definition of Martensite.—A practical definition of martensite is as follows:[1] A microconstituent or structure in quenched steel characterized by an acicular or needlelike pattern. It has the maximum hardness of any of the decomposition products of austenite.

TYPICAL QUENCHING DIAGRAMS

A convenient means of summarizing the data obtained by quenching a given steel at different rates of speed is shown in Fig. 6-X. In this diagram the changes in the transformation temperatures with cooling rates are correlated with the structures. The correlation between this type of diagram and the series of cooling curves, previously discussed, is easily seen. The upper transformation curves show the gradual lowering of the *Ar* with increased cooling rates and has been designated *Ar′*. The corresponding structures with increasing rates of cooling are coarse pearlite, fine pearlite,[2] and very fine pearlite. At a given rate of

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[1] A.S.M. "Metals Handbook," p. 8, Cleveland, Ohio, 1939.

[2] Fine pearlite may be called *sorbitic* pearlite, and very fine pearlite may be termed *primary troostite*. The terms "sorbite" and "troostite" are used

cooling there is a split in the temperature of transformation, so that it takes place partly at high, and partly at low temperature, and results in the formation of the mixed structure. The low transformation temperature, denoting the formation of martensite, is designated *Ar''*. The cooling rate at which the *Ar'* disappears is the critical cooling rate and marks the disappearance of very fine pearlite and the retention of martensite or a martensitic-austenitic structure.

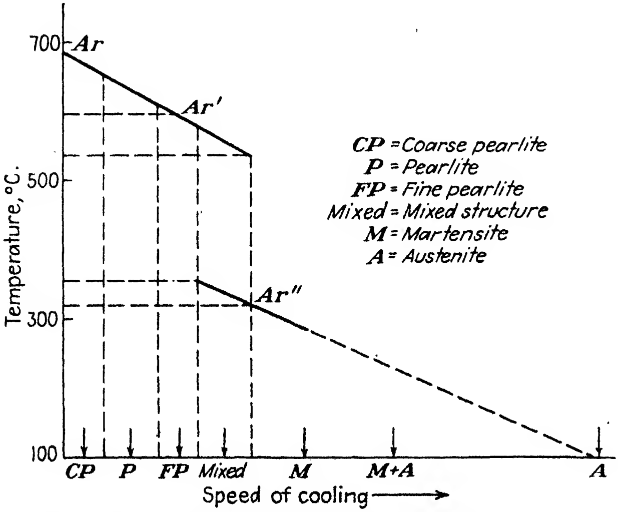

FIG. 6-X.—Rate of cooling versus critical point and microstructure.

Considerable discussion has arisen over the nomenclature that should apply to the various states of fineness or dispersion of the lamellar aggregate. As is shown in the discussion, when the rate of cooling from above the critical is increased, the cementite lamellae becomes progressively finer and finer. Consequently, this increased state of fineness destroys the normal structural proportions of approximately 7 parts of ferrite to 1 part of cementite. Thus, the point of issue is whether the term "pearlite" should apply to all the various states of dispersion of the cementite lamellae, irrespective of the structural proportion

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in this book to denote the temper structures, while the above mentioned terms are used to differentiate the temper structures from those formed by controlled cooling rates.

of free ferrite to cementite, or only to the aggregate formed by slow cooling and to which the original name "pearlite" was given.

Davenport and Bain,[1] in their classical investigations, have determined the rate of transformation of austenite at various constant temperatures below the equilibrium range and have summed up their results for a number of carbon steels on time-temperature charts. Figure 7-X is a typical illustration of the trend taken by the steels studied to form *S*-type curves. This time (logarithmic)-temperature curve shows the results of quenching a group of small specimens of eutectoid character from above 730°C. (1346°F.), the lowest temperature of austenite stability, into a quenching bath of molten lead held at a chosen constant temperature. The specimens were held in the bath for given intervals of time and, at the end of that time, withdrawn and quenched in an iced brine. Subsequent microscopic examination revealed the extent of decomposition of the austenite during the time held at the constant temperature of the quenching bath. The rate of decomposition of the austenite was thus measured.

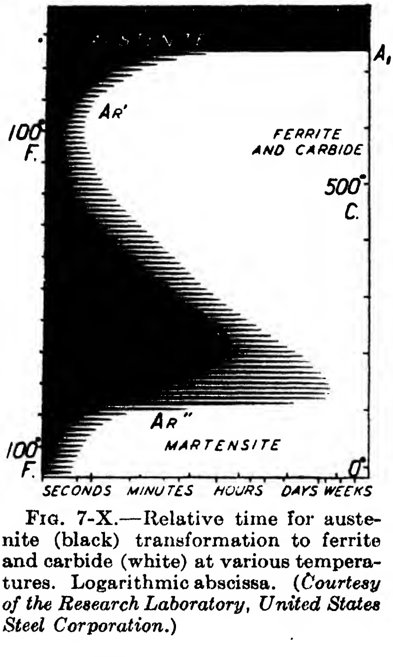

Fig. 7-X.—Relative time for austenite (black) transformation to ferrite and carbide (white) at various temperatures. Logarithmic abscissa. (*Courtesy of the Research Laboratory, United States Steel Corporation.*)

The diagram in Fig. 7-X should be visualized as having three time zones at any temperature level below the characteristic equilibrium temperature A_E. These three zones are as follows: (1) the zone in which the austenite has not yet started to transform, even though it is unstable; this zone is that portion of the

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[1] Davenport and Bain, Transformation of Austenite at Constant Subcritical Temperatures, *Trans. A.I.M.M.E.*, **90**, 117–154 (1930). Bain, E. C., On the Rates of Reaction in Solid Steel, *Trans. A.I.M.M.E.*, **100**, 13–46 (1932) and Factors Affecting the Inherent Hardenability of Steel, *Trans. A.S.S.T.*, **20**, 385–428 (1932).

curve shown entirely in black, with the exception of the black portion above A_1, in which area the austenite is entirely stable; (2) the zone in which the transformation proceeds, at constant temperature, to completion. This is shown by the shaded portion of the curve. We can thus visualize that there is a whole series of *S*-shaped curves within this shaded area, the area being bounded on one side by a curve that just outlines the black area and represents, shall we say, practically zero transformation, and on the other side by a curve that just outlines the white area and represents complete transformation. Between these two extremes there may be other curves representing different amounts of transformation for a given time at any particular temperature level; (3) the zone in which the transformation is complete and the structure and properties of the treated steel are definitely set. This area covers the portion of the curve that is entirely white.

The diagram also shows that there are two temperature ranges of relatively rapid transformation, one in the vicinity of 1000–1100°F. and the other at approximately room temperature; the time, in both cases, for complete transformation is a matter of seconds. It appears that the level of rapid transformation at 1000 to 1100°F. is not more than about 300° below the equilibrium point Ac_1 for all the steels. There are also two temperature ranges of relatively slow transformation, one at high temperatures just under the equilibrium line A_E and the other in the range of 900 to 300°F., the time in both these cases ranging from approximately one minute to a matter of days and even weeks.

The position of the curve with respect to the time-temperature coordinates is dependent upon the composition of the steel and the austenitic grain size. The general trend of an increase in alloy content or austenitic grain size, or both is to shift the curve to the right of the diagram. These factors will be discussed under the appropriate sections. With respect to the usual S.A.E. carbon steels, it is found that the curves will shift to the right with increasing carbon content. The very low carbon steels have the nose of the high-temperature, rapid-transformation zone so close to the temperature coordinate that it is practically impossible to quench rapidly enough to suppress the transformation so that full hardening may be obtained.

Fig. 8-X.—7, nodular fine pearlite associated with martensite in a quenched 0.85 per cent carbon steel. × 2,500. 8, first traces of pearlite formation after 320 sec. at 1300°F. (705°C.) in an 0.85 per cent carbon steel. × 1,000. (*Courtesy of American Society for Metals.*) (*See footnote, page* 219.)

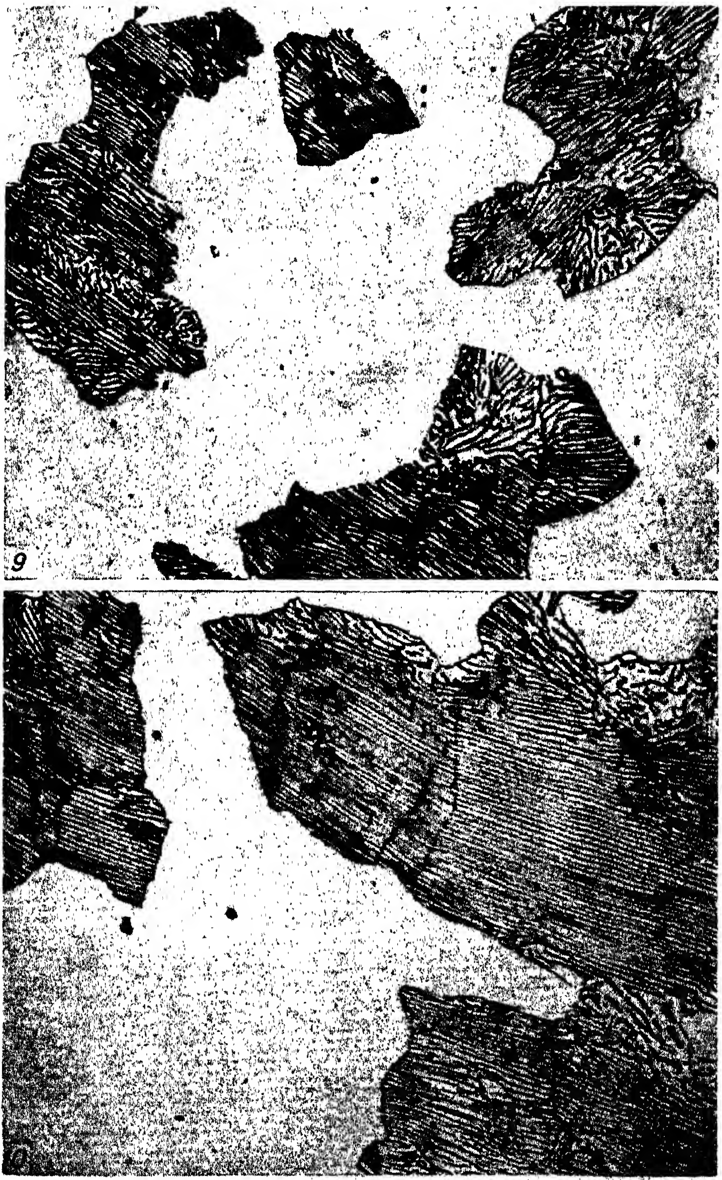

FIG. 9-X.—9, pearlite formation in steel of Fig. 8-X (8) advanced to 25 per cent (75 per cent austenite) after 1,150 sec. × 1,000. 10, pearlite formation as in Figs. 8-X(8) and 9-X(9) advanced to 50 per cent (50 per cent austenite) after 1,320 sec. × 1,000. *(Courtesy of American Society for Metals.)* (S

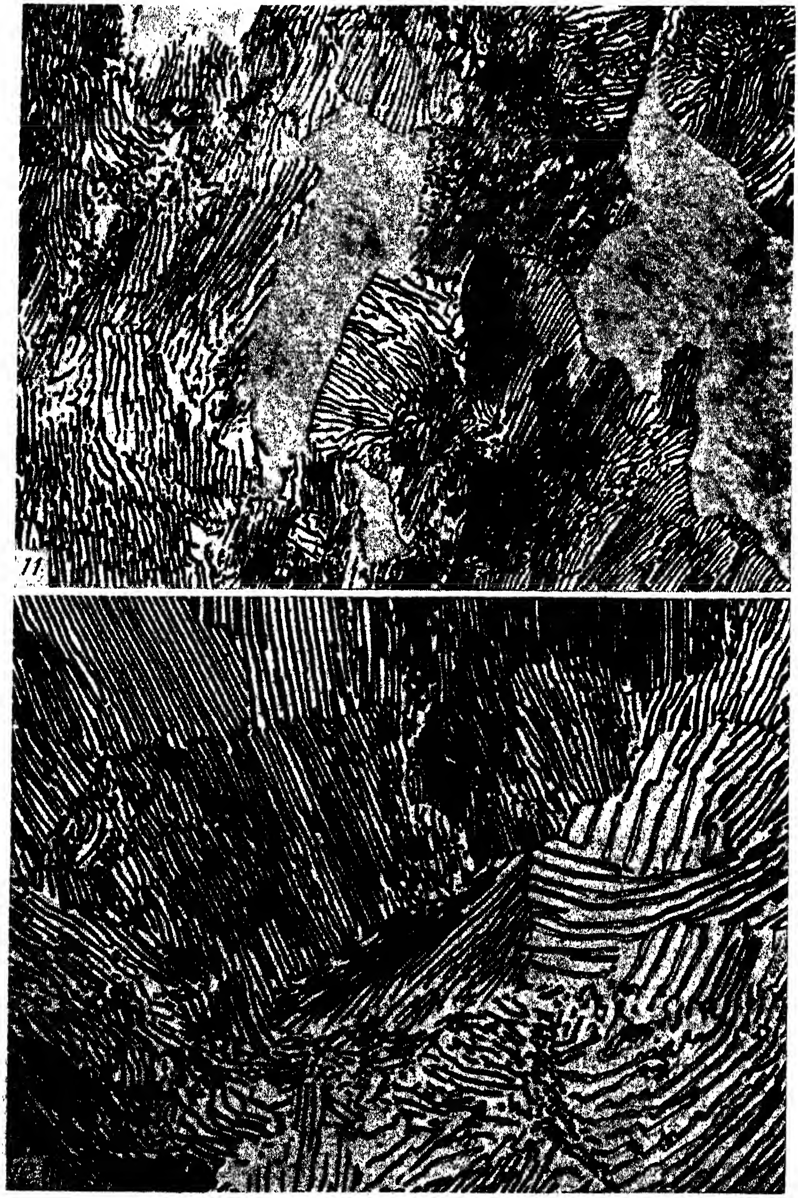

FIG. 10-X.—11, pearlite formation as in Figs. 8-X(8), 9-X(9), and 9-X(10) advanced to 75 per cent, (25 per cent austenite) after 1,450 sec. × 1,000. 12, pearlite formation completed in steel of Figs. 8-X(8), 9-X(9), 9-X(10), and 10-X(11) after 4,000 sec. × 1,000. (*Courtesy of American Society for Metals.*)

The rate of transformation of austenite at a given temperature level is shown as follows: Figs. 8-X (8), 9-X (9) and (10) and 10-X (11) and (12) show the progress of decomposition of the austenite in an 0.85 per cent carbon steel at 1300°F., after 320, 1,150, 1,320, 1,450, and 4,000 sec., respectively, to pearlite. The actual background of these figures is not austenite but martensite, which has resulted from the quench applied to halt the reaction at the desired stage in its progress.

The connection between the S-shaped curve and the conditions represented by the cooling curves (Fig. 1-X) should now be evident.[1] The S-shaped curve sums up the rate of transformation with respect to time and temperature. It shows that at the high temperatures of transformation, the transformation product consists of coarse lamellae of ferrite and carbide; at progressively lower temperatures the pearlite becomes progressively finer until, at about 1000°F., when the transformation is most rapid, it becomes so fine that it is practically unresolvable even with a lens of the highest power. If the transformation takes place at very low temperatures, below about 300°F. (150°C.) the reaction is a different one and is rapid. The carbide is not precipitated but is retained as the unstable, supersaturated solid-solution, martensite.

The photographs in Figs. 2-X, 3-X, 4-X, and 5-X illustrate the typical microstructures of the product of austenite transformation at various arbitrarily chosen temperature levels, namely, 1325°, 1300°, 1250°, 1200°, 1100°, 1000°, and 100°F.

The rosettelike structure shown in Fig. 8-X (7) is termed *nodular fine pearlite* or *primary troostite.* This lamellar product is the result of the constant-temperature transformation between the temperatures of about 950 to 1150°F. and represents the normal limit of fineness of carbide lamellae.

At transformation temperatures of between about 950 to 300°F. a new constituent appears, which has been given various names by different observers, such as "intermediate structure," "pseudomartensite," "acicular troostite," "acicular constituent," and "Bainite." There are two distinct forms of this constituent,

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[1] It should be noted that the curve for the starting of the transformation with falling temperature is not strictly the same as under the conditions shown in Fig. 6-X, which is for a series of constant-temperature levels. The difference, however, is slight and can generally be neglected in practice.

FIG. 11-X.—Acicular structure characteristic of transformation that occurs just below the knee of the *S* curve. Incompleted reaction in eutectoid steel at 1000°F. (540°C.). × 1,500. (*See footnote, page* 219.)

FIG. 12-X.—Acicular structure characteristic of transformation near the martensite temperature range. Incompleted reaction in eutectoid steel at 550°F. (290°C.). × 2,500. (*See footnote, page* 219.)

the high-temperature form and the low-temperature form, with intermediate gradations of structures between these two. The appearance of the constituent is characteristically feathery and acicular at both high and low transformation temperatures, as is shown in Figs. 11-X[1] and 12-X.[1] The exact nature of the constituent is as yet somewhat problematical, although it is readily distinguishable from true martensite by its etching characteristics, its appearance, and its softness.

TEMPERING

The hardening treatment, discussed in the previous section, will impart to the steel (1) maximum hardness, (2) maximum tensile strength, (3) minimum ductility, and will also produce an unstable, internally stressed condition.

In this condition, the steel, except in a few cases, could not have a very wide commercial application. By proper application of heat, however, it is possible to obtain any desired combination of physical properties within the capacity of that particular steel; in other words, the hardness and tensile strength can be reduced, the ductility increased and the unstable or stressed condition relieved.

The result of reheating or tempering a fully quenched steel is as follows:

Martensite, as it has been defined, is a highly supersaturated solid solution of carbon or carbide in ferrite or alpha iron that has precipitated along certain favored crystallographic planes. Steel in this condition is obviously in an unstable condition. With the application of heat, the carbon, which is held in solid solution over and above that which alpha iron is ordinarily capable of holding, will have increased opportunity to migrate and diffuse; the result will be its combination with iron and a shower of carbide particles of extreme fineness. As the temperature is raised there is increased opportunity for the migration and coalescence of these fine carbide particles into larger and larger particles, until the critical temperature is reached, at which point there is maximum coalescence. Above this temperature there would be, of course, transformation to austenite. It must be emphasized that in no case will the reheating of a hardened steel

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[1] VILELLA, GUELLICH, and BAIN "On Naming the Aggregate Constituents in Steel," *Trans. A.S.M.*, Vol. XXIV, No. 2, p. 225 (1936).

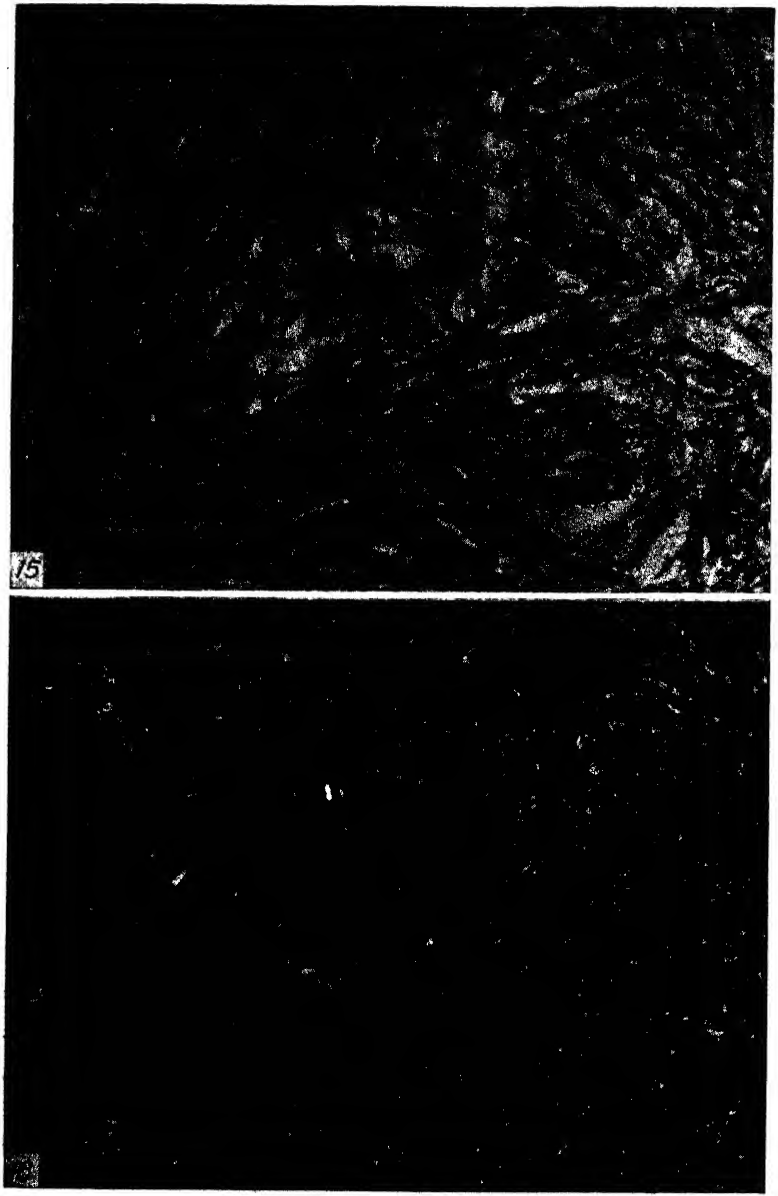

FIG. 13-X.—15, typical martensite in eutectoid steel, 0.85 per cent carbon. × 2,000. 16, tempered martensite, developed by heating material shown in 15 for 2 hr. at 445°F. (230°C.). × 2,000. (*Courtesy of American Society for Metals.*) (*See footnote, page* 219.)

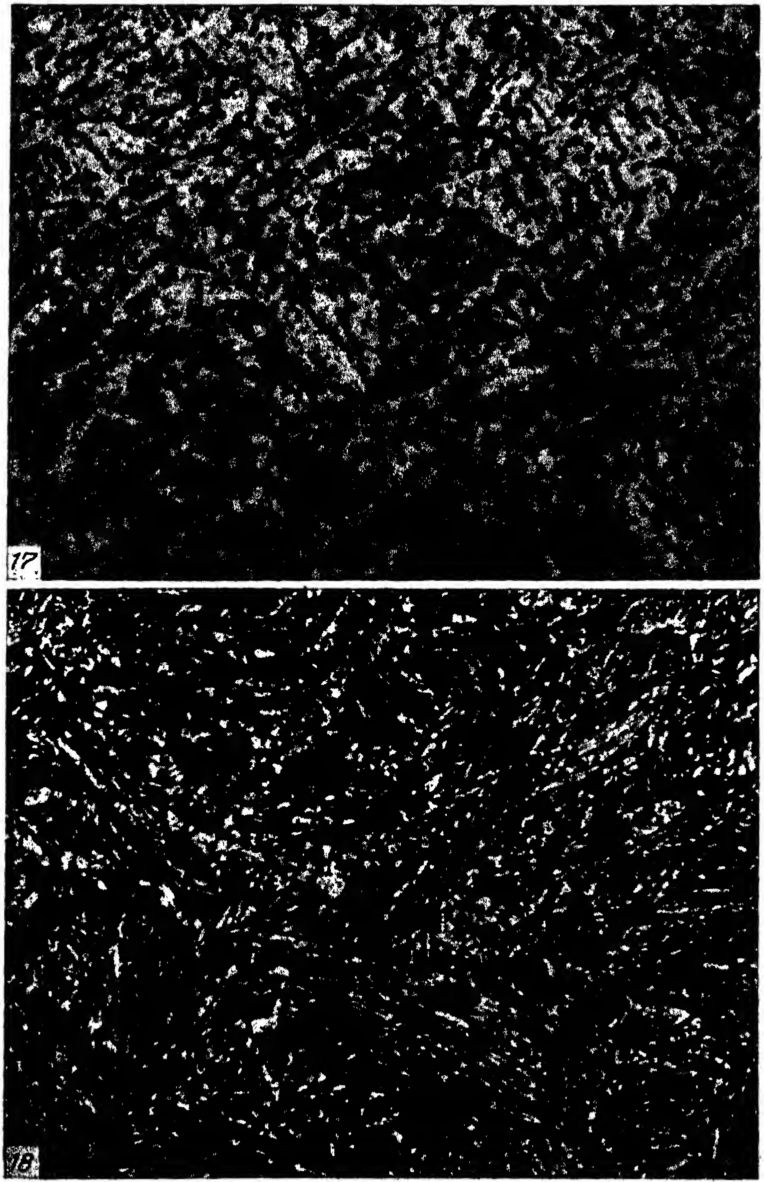

Fig. 14-X.—17, tempered martensite developed by heating material of Fig. 13-X(15) for 5 min. at 1200°F. (650°C.). × 2,000. 18, tempered martensite prepared by heating material of Fig. 13-X(15) for 2 hr. at 840°F. (450°C.) to develop a hardness of 41 Rockwell C. × 2,000. (*Courtesy of American Society for Metals.*) (*See footnote, page* 219.)

result in a lamellar structure; it will always be globular in character.

Photomicrographs shown in Figs. 13-X (16), 14-X (17) and (18), respectively, show the structures resulting on tempering martensite of Fig. 13-X (15). On the reheating of martensite to a temperature of from 300 to 750°F. (150 to 400°C.) there is formed a softer and tougher structure, which is known as *troostite*. On the continued heating of troostite or martensite to a temperature of from 750 to 1290°F. (400 to 700°C.) there is formed a structure that still retains high strength, although less than troostite, but has much greater ductility. It is known as *sorbite*. Then on the heating of martenstite, troostite, or sorbite to a temperature of about 1290°F. (700°C.), but below the A_1 critical, there is formed spheroidized cementite in ferrite (Fig. 4-IX). This spheroidal condition may also be formed by reheating fine pearlite at temperatures above its formation temperature, *e.g.*, 1100°F. (600°C.), to just below the A_1 critical. If this tempering should take place at 1200°F. (650°C.), we will perhaps get a very

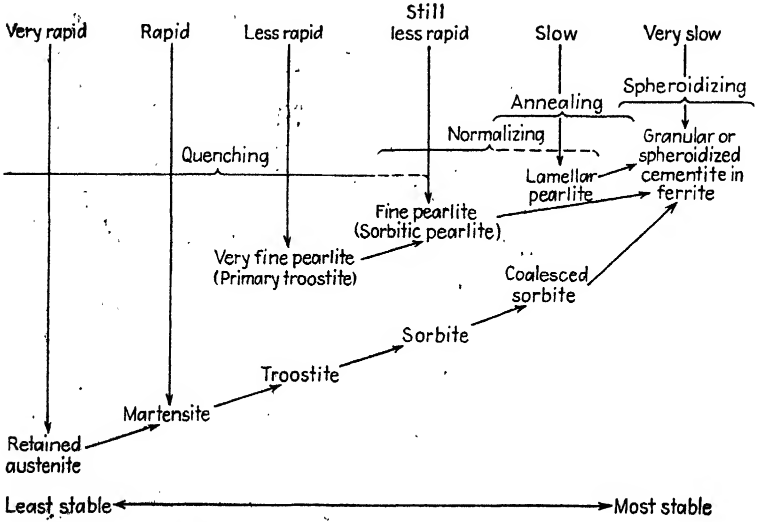

Fig. 15-X[1].—Diagram illustrating how various structures may be obtained in a eutectoid steel by varying the rate of cooling and of reheating after quenching. (*After A. A. Bates.*)

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[1] Bates, A. A., "Fundamentals of Ferrous Metallurgy," A.S.M., Lecture Course, Lecture VII.

coarse sorbitic structure. A very coarse lamellar structure will coalesce with the spheroidal structure very slowly at any tempering temperature, which would incidentally be quite high, owing to the high temperature of the formation of coarse pearlite. This coarse spheroidal structure has, as previously noted, maximum ductility and low tensile strength.

Figure 15-X shows diagrammatically how the various structures that have been discussed may be obtained in a eutectoid steel by varying the rate of cooling and of reheating after quenching.

Physical Properties as Affected by Tempering.—The general trend of the effects of tempering may be studied from the results in Table 2-X, which give the increase in toughness and decrease in hardness in a 0.43 per cent carbon steel that has been fully hardened by quenching in water and then reheated to the successively higher temperatures shown.

TABLE 2-X

Temperatures	Rockwell hardness, C-scale	Charpy impact, ft.-lb.
As quenched	60	3
Tempered at 400°F	56	7
Tempered at 600°F	50	5
Tempered at 800°F	42	12
Tempered at 1000°F	34	32
Tempered at 1200°F	24	56

The regular decrease in hardness will also be accompanied by a regular decrease in the tensile properties. Typical changes in tensile strength, yield strength, elongation, reduction of area, and hardness for the S.A.E. steels and others can be found in charts of physical properties carried in societies' publications and manufacturers' handbooks.

Along with the change in physical properties mentioned there is an increase in the impact toughness. As the tempering temperature is raised, the impact toughness gradually increases, particularly beyond the temperature of about 800°F.

In the region between about 450 to 650°F., however, there is a marked decrease in the toughness on tempering. The decrease in toughness within this range is probably due to the decomposition to alpha iron and carbide or carbon of a certain small proportion of austenite that is always retained when the steel is quenched to martensite. The decomposition does not start until a temperature of about 450°F. is reached, and it becomes complete at about 650°F. The product of this decomposition, distributed throughout the structure, tends to bring about brittleness in impact. There is no apparent change in the hardness, probably because of the small proportion of the products present, but in some steels the hardness increase is quite apparent.

CONVENTIONAL HEAT-TREATING OF STEEL

In ordinary practice the transformations are produced while the piece is being cooled at some desired rate and not by causing the transformation to take place at a constant temperature. The curves that have just been studied will, however, indicate what is occurring under ordinary practice. This correlation is illustrated in the diagram of Fig. 16-X, which shows, first, the transformation-rate curves for the beginning and end of transformation of a medium carbon steel and, second, the relationship between cooling rate and the isothermal *S*-curve. In the cooling curve *A*, which shows a very slow rate of cooling, as in annealing, the bar of steel will on reaching temperature t^1 start to transform to pearlite. As the temperature drops, the rate at which pearlite forms increases, and when the point t^2 is reached on the curve the pearlite transformation is complete. Curves *B* and *C* show increased rates of cooling, somewhat analogous to air cooling or normalizing. The pearlite formed in each of these three cases will vary from coarse in the case of curve *A* to fine in the case of curve *C*. The hardness of the structure will vary according to the lamellar spacing, the annealed structure being the softest.

With still further increase in the cooling rate, as in curve *E*, which illustrates a rapid quenching operation, such as quenching in a bath of oil or water at room temperature, no transformation will take place until the temperature reaches t^3, at which point martensite forms. In this case, the critical cooling rate has been exceeded, so that there has been no opportunity for the rapid pearlite transformation, at 1000°F., to take place.

At cooling rates intermediate between *C* and *E*, as in curve *D*, the final structures will consist of a certain proportion of both pearlite and martensite. In curve *D* the transformation will start at t^5 but since the time at the temperature is insufficient to allow complete transformation the pearlite will stop forming at t^6 with the remaining austenite cooling to a lower temperature t^7, where it transforms to martensite. The resultant structure will, therefore, be a mixture of pearlite and martensite. The pearlite reaction under these conditions will take place in the boundaries of the austenite grains and will appear as the dark etching nodules of pearlite previously discussed (they may appear within the grains).

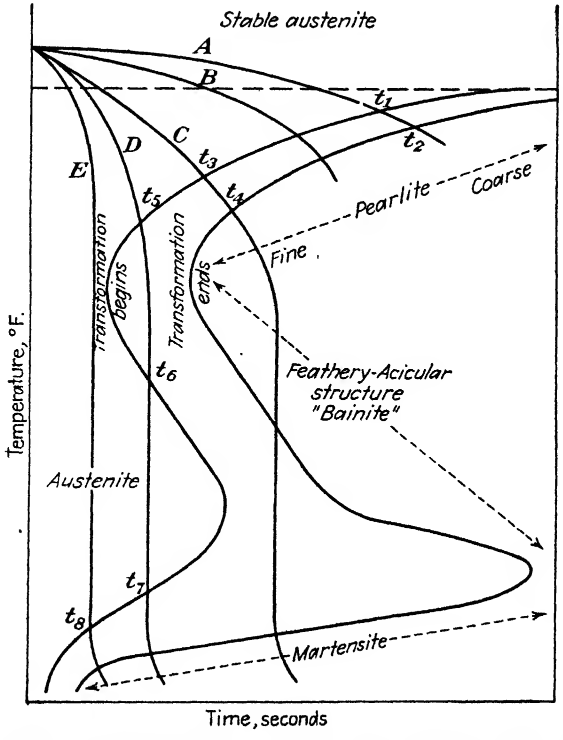

FIG. 16-X.—A quenching diagram for a bar of medium carbon steel.

In the conventional cooling practices, the conditions that bring about the transformation of the products known as *Bainite*, *i.e.*, constant temperature transformation in the range of 950 to 300°F., will not be encountered, as quenching baths are seldom or never at 300 to 700°F.

AUSTEMPERING PROCESS FOR HEAT-TREATMENT OF STEEL

In practice[1] this method for heat-treatment of steels is confined to those constant-temperature levels that produce all or almost all Bainite in the transformed steel. It consists in heating the steel until it is completely austenitic, then quenching into a molten salt or low-melting alloy bath held accurately at some predetermined temperature for a definite period of time known to be sufficient for complete transformation to Bainite.

In this process, as well as in the conventional process, such factors as composition, mass, and the austenitic grain size of the steel, and its cooling rate from the hardening temperature influence the process and the result.

The successful application of austempering to any steel depends upon a thorough knowledge of its transformation characteristics. This means that all or part of the *S*-curve must be known, because such factors as composition and austenitic grain size may considerably shift its location on the time axis and have a large effect on the time interval at the temperature used. Since one of the reasons for the application of this process is to obtain high hardness, it is recommended that the carbon content in the plain carbon steels be not less than 0.60 per cent.

Success of the process not only depends on a knowledge of the curve characteristics but also on the cooling rate of the entire section being treated. Thus, in order to produce the structure and ultimately the hardness desired, the section must be cooled at such a rate, from its austenitic range, that no transformation takes place until the desired temperature is reached. For complete transformation of the entire section to the desired structure every part must be cooled at this rate, which corresponds to the rate illustrated by curve *E* in Fig. 16-X. When the section size is beyond a certain critical size for a given composition of steel, complete transformation to hard, tough Bainite is no longer possible, as the rate of cooling (as the rate in the center of a large section) will be retarded and the transformation product will be pearlite. This rate might correspond to curve *D* in Fig. 16-X. In the plain carbon steels the treatment is successful in varying sizes, depending upon the analysis of carbon and manganese.

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[1] Legge, E. E., The Industrial Application of Austempering, *Metals and Alloys*, **10**, 228–242 (1939).

For example, with a carbon content of 0.95 to 1.05 per cent and manganese 0.30 to 0.50 per cent, the suggested maximum austemperable size is around 0.148 in. diameter; with the same carbon with 0.60 to 0.90 per cent manganese, the size is 0.187 in. diameter. It is possible to increase the size that can successfully be treated, by the addition of alloying agents, to as much as 1.0 in. or more.

Physical Properties[1] as Affected by the Quench and Temper Method versus the Austempering Method.—Table 3-X shows the results of mechanical property tests made on 0.180 in. diameter steel rods, heat-treated by the quench and temper method and by austempering to the same hardness.

TABLE 3-X.—MECHANICAL PROPERTIES OF STEEL
(Average of three tests)
0.74 per cent C, 0.37 per cent Mn, 0.145 per cent Si, 0.039 per cent S, 0.044 per cent P

Property	Quench and temper method*	Austempering*
Rockwell C	49.8	50.1
Tensile strength, p.s.i.	259,000	259,300
Elongation in 2 in., per cent	3.75	5.0
Reduction of area, per cent	26.1	46.4
Impact, ft.-lb.†	14.0	36.6

* Heat-treatment:

Quench and Temper	Austempered
Pb bath 1450°F.—5 min.	Pb bath 1450°F.—5 min.
Oil quench	Quench Pb—Bi bath 600°F.—20 min.
Tempered 650°F.—30 min.	Quench in water

† Foot-pounds absorbed in breaking 0.180 in. round, unnotched specimens.

In general, it is found that the tensile strengths obtained by austempering to a given hardness are of the same order as those resulting from heat-treating in the regular way to the same hardness.

The improvement in elongation for equivalent tensile strength by austempering may be as high as 30 per cent in high hardnesses, but in low hardnesses it is little, if any, superior to the conven-

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[1] LEGGE, E. E., The Industrial Application of Austempering, *Metals and Alloys*, **10**, 228–242 (1939).

tional method. Austempering, however, brings a very large improvement in the reduction of area at the higher hardness values.

The impact strength, as tested on unnotched, round specimens, shows a large improvement by austempering, particularly for steels treated within the hardness range of 48 to 55 Rockwell C, where austempering is most advantageous.

It must be noted, in appraising this method, that there are definite hardness ranges in which the austempered steels are considerably superior to those quenched and tempered to the same hardnesses, and other ranges in which the austempering offers little, if any, advantage. The better properties of austempered steels are found in the hardness range of from 48 to 58 Rockwell C.

Martempering.[1]—This type of heat-treatment, although it is discussed under austempering, differs basically in the end results. The operation consists in cooling rapidly to just above the Ar_2 martensite critical, equalizing the temperature throughout the section, and then slowly cooling through the Ar_2 to produce martensite. The conventional tempering operation should immediately follow the above treatment.

The operation involves an accurate time quench in a medium that will cool the section, even at its center, past the rapid pearlite reaction at the nose of the *S*-curve to a temperature just above the martensite critical for the steel. After being held at this temperature to allow the establishment of temperature equilibrium, the piece is removed from the cooling medium (molten salt is found to be satisfactory) and cooled in the air. The austenite that is so stable at the equalizing temperature transforms to martensite on air cooling at a fairly uniform rate throughout the section.

In the usual full-quenching practice, there is always danger of distortion and cracking as a result of several natural phenomena: (1) The steel contracts as it cools, (2) it expands when changing from austenite to martensite, and (3) these changes occur at different places, since the surface is cooler than the center. The advantage of the martempering treatment over the conventional procedure lies, therefore, in the fact that it minimizes the residual

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[1] SHEPHERD, B. F., *Iron Age*, Jan. 28 and Feb. 4, 1943; and *Metal Progress*, July, 1943, p. 105.

stresses and strains and provides greater freedom from distortion and hardening cracks.

HARDENABILITY

In the practical quenching operations just discussed another very interesting phenomenon can be observed. Figure 17-X, which illustrates this phenomenon, shows the polished and etched cross section of a 1-in. round of eutectoid carbon steel quenched from above the critical range into brine. Hardness determinations were made as shown on the diameters of the specimen with a Rockwell machine using the C-scale. It is quite apparent from this figure that there are at least two zones of transformation, as is evidenced by the differential etching characteristics, and that each zone has a different hardness, as is shown by the depth of the Rockwell impressions. In the light of our previous study, the reason for this effect should be apparent. At the outside there is a narrow case of metal that has cooled fast enough to get past the rapid transformation zone at 1000°F. and has transformed at the quenching-bath temperature (150 to 200°F.) to martensite, with a hardness of C-60 or better. The cooling characteristics of this zone will follow the curve *E* in Fig. 16-X.

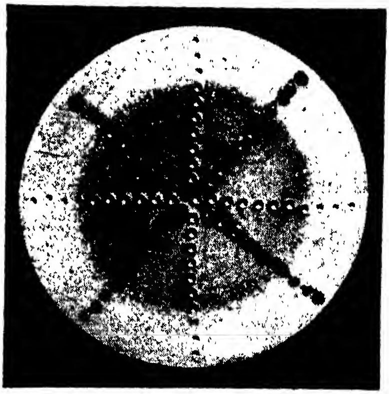

FIG. 17-X.—Section of 1-in. round eutectoid carbon steel quenched in brine. Hardness determinations were made as shown on diameters of specimen. Note change in penetration at circle representing the critical cooling rate; outside martensite, inside fine pearlite. (*Courtesy of Research Laboratory, United States Steel Corporation.*)

In the center of the round the cooling rate will be very much slower, so that the transformation will be to pearlite with a hardness of approximately C-40. The cooling characteristics of this zone will follow after curve *B* or *C* in Fig. 16-X.

Somewhat below the surface and between the two previously discussed zones the round will be subjected to an intermediate cooling rate, so that the transformation will occur partly at high temperatures and partly at low. The resultant microstructure will be very fine pearlite (nodular) and martensite, having a compromise hardness between C-42 and C-60, depending upon the relative amounts of each constituent present. The cooling characteristics of this zone will follow curve *D* in Fig. 16-X.

The hardness results of such quenching operations can be plotted in the form of a curve in which hardness is plotted vertically and the diameter of the specimen, horizontally (Fig. 18-X). From a curve of this type, differences in actual hardness or in hardness penetration can easily be determined.

The term "hardenability" is used to express the factors just discussed. This term for a given heat of steel should be described[1] as meaning both the ability to take on surface hardness on quenching and the ability to depth-harden; however, for practical purposes it refers to the depth of hardening. In a more inclusive sense, it may be considered as the relationship between the rates of cooling from above the critical range and the hardnesses and microstructure developed as a result.

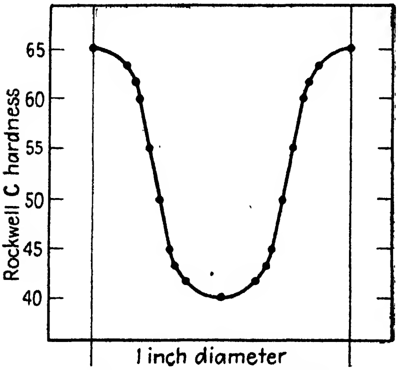

FIG. 18-X.—Typical hardness distribution curve for a quenched 1-in. round of a given steel.

Prior to the discussion of the details of hardenability it is essential to have a general picture of the difference between hardenability and the potential maximum hardness. For example, comparing an S.A.E. 1075 steel with a 1035, it may be found that the former in the hardened condition has a hardness of C-65, while the latter has a hardness of C-58, yet the two steels may have the same hardenability, *i.e.*, they may both require the same rate of quenching to bring about their respective full hardnesses.

It has been found that hardenability is an inverse function of the critical cooling rate.[2] This means that, if a steel requires

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[1] BURNS, MOORE and ARCHER, Quantitative Hardenability, *Trans. A.S.M.*, **26**, 1–36 (1938).

[2] FRENCH, H. J., A Study of the Quenching of Steels, *Trans. A.S.S.T.*, **17**, 646–730 (1930); DAVENPORT and BAIN, Transformation of Austenite at Constant Subcritical Temperatures, *Trans. A.I.M.M.E.*, **90**, 117–154 (1930).

a rapid quench to harden, it is referred to as having low hardenability; while, should the required quench not be so drastic, the steel has higher hardenability. Thus, a ¾-in. round of plain carbon steel must be drastically quenched in water to produce full hardening and it is said to have low hardenability, while a ¾-in. round of steel with sufficient alloy will harden in the water quench as well as in a milder oil quench and is said to have high hardenability. This difference is also noted when, quenching under like conditions, the steel of high hardenability will harden to a greater depth.

Regarding the practice of hardenability control to obtain the desirable characteristics in the finished part, this is the general rule:[1] Low hardenability for toughness, high hardenability for strength and for ease of hardening.

Hardness.—The hardness of a steel can be expressed quantitatively by values as measured by any of the regular methods of hardness testing. The maximum hardness attained,[2] as was previously pointed out, is entirely dependent upon the carbon content, regardless of alloys in the amounts present in the common S.A.E. steels, and regardless of the austenitic grain size. The attainment of this hardness is dependent upon the following conditions: (1) that all the carbon is in solution at the time of quenching, (2) that the critical cooling rate is reached or exceeded at the time of quenching, and (3) that no appreciable austenite is retained after the quench.

It should be emphasized that, in order to attain the maximum hardness that a given carbon steel can potentially produce, the quench may be impractical. For example, it is possible to obtain Rockwell C-48 with a 0.20 per cent carbon steel, but the piece must be very thin and quenched in iced brine. A 0.20 per cent carbon-alloy steel, however, can be easily hardened to the above hardness level. With larger sizes it is impossible to attain the maximum potential hardness.

Depth of Hardening.—The penetration, or depth of hardening, for a given steel can be measured by a number of methods: (1) fundamental hardenability (transformation rates at constant

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[1] GROSSMANN, M. A., Hardenability, as It Affects Heat Treated Parts, *Metal Progress*, **33**, 373–376 (1938).

[2] BURNS, MOORE and ARCHER, Quantitative Hardenability, *Trans. A.S.M.*, **26**, 1–36 (1938).

temperature); (2) critical cooling rate; (3) hardness gradient (hardness distribution curves); (4) Jominy, G.M.C., or end-quench test; (5) fracture-and-etch test (Shepherd standard); (6) microstructure.

Fundamental Hardenability.—The most complete and thorough knowledge of the hardenability of a heat of steel is obtained by the study of its reaction rates. This has been discussed in detail in the previous section. The difficulty with this method is that it is long and laborious and could not be applied to commercial control.

Critical Cooling Rate.—The determination of the critical cooling rate (the rate through the transformation range that must be equalled or exceeded to produce complete hardening to martensite) of a given section of steel in a particular quenching medium is also a means of measuring the hardenability. The procedure would be a matter of determining the cooling rates from the time the steel enters the critical range until it has passed the 1000°F. zone. This would indicate whether the steel has formed any pearlite during quenching and what portion that has not changed to pearlite will change to martensite. The difficulty with this method is that it involves a complicated experimental procedure.

Hardness Gradient.—One of the most common and accurate methods for describing hardenability is by showing the hardness across a quenched cross section. The method employed is to quench a round bar under selected conditions, which are governed by the investigation or the practice to be employed, to cut or break it in half after quenching, and then to grind it to a smooth surface. Next, the hardness readings are taken across one or more of the diameters. The readings taken are averaged and a curve is plotted, such as the one shown in Fig. 18-X. A comparison of the values determined by this type of hardness traverse on a bar of a given size with others of the same size but of different analyses will show their differences in hardenability. It is also possible to check bars of various diameters of identical analysis by this same procedure.

Hardness Distribution.—The hardness distribution curves for a series of bars of sizes ranging from ½-in. round to 5-in. round, for an S.A.E. 1045 steel quenched in water and oil, are shown in Figs.

19-X and 20-X.[1] Such progressive hardnesses indicate that the sections with high hardness are martensitic, those of intermediate hardness are mixtures of martensite and pearlite, and those of low hardness are pearlite of varying structural coarseness.

The same steel and section sizes quenched in oil show a very striking difference in hardness values from those subjected to the water quench. The oil quench, which is much less drastic than water (approximately one-third as fast as water), produces a surface hardness of only Rockwell C-29 in the smallest size, with a center hardness slightly under this figure. Both surface and center hardnesses progressively decrease from these values as the section size increases—an indication that this steel, having

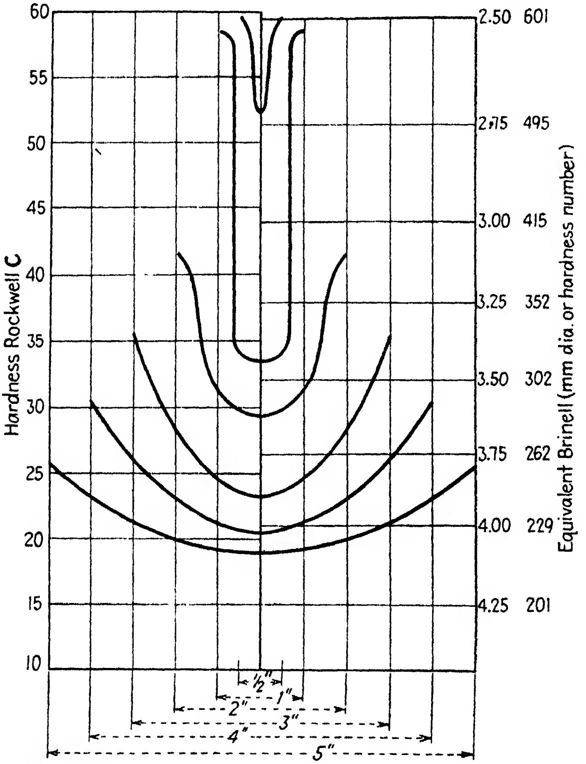

FIG. 19-X.—Hardness distribution in various sizes of quenched round bars. 1045 steel quenched in water.

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[1] U.S.S. Carilloy Steels, Carnegie-Illinois Steel Corp., Pittsburgh, 1938.

low hardenability, fails to harden when it is oil quenched and will in all sizes transform to soft pearlite.

This method of determining hardenability can be used on plain carbon or alloy steels of deep- or shallow-hardening grades. It is not suitable for measuring steels in the carburized condition. The procedure of this test, when duplication of results is attempted, involves difficulties in controlling the severity of the quench, differences in heating, differences in surface finish, and the dangers of drawing the hardness during sectioning. This method is not particularly suitable for the deep-hardening alloys unless very large sizes are used, and then there is no limit to the successful application, other than the problem of sectioning the hardened bar.

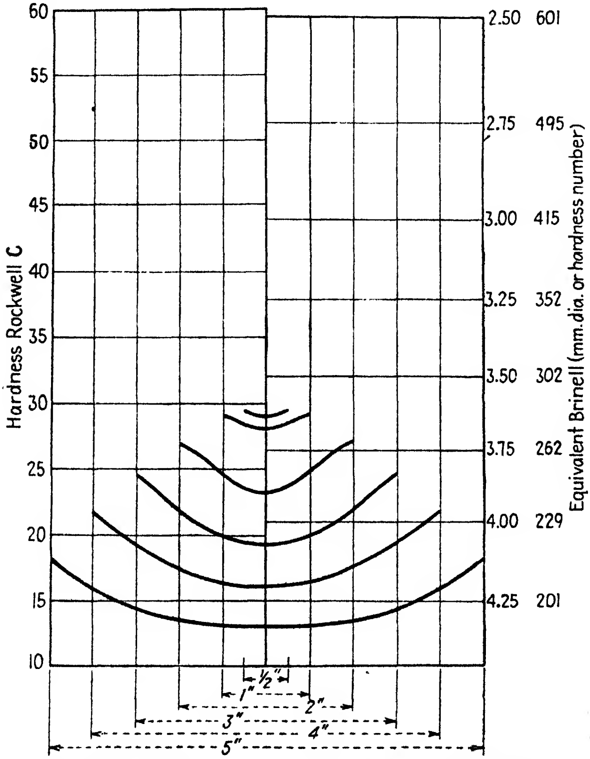

FIG. 20-X.—Hardness distribution in various sizes of quenched round bars. 1045 steel quenched in oil.

Jominy, G.M.C., or End-quench Test.—This method of measuring hardenability[1] consists in heating a standard end-quench test bar, a 1-in. round, 3⅞ in. long, that has been previously normalized or annealed to a hardening temperature and then placing it in a fixture so that the 1-in. round face is over a water outlet. The jet of water is then turned on by a valve and the face is quenched (Fig. 21-X, *A*).[2] After the specimen is cooled, it is prepared for hardness reading by grinding longitudinal flats on opposite sides. Rockwell C hardness readings are made over the length of the bar at 1/16-in. intervals, and these values are recorded by plotting them against the distance from the quenched end on a chart with a curve resulting as in (Fig. 21-X, *B*).[2]

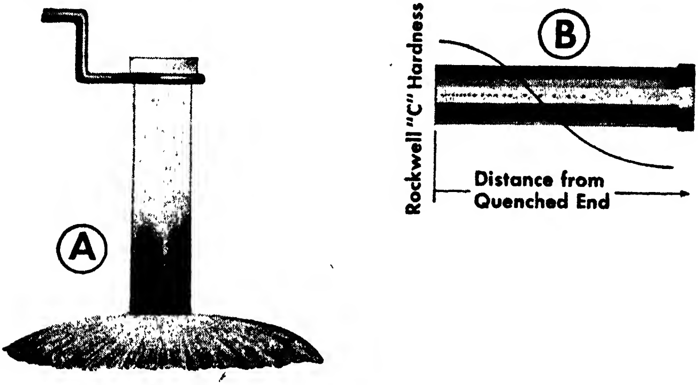

FIG. 21-X.—End-quenched hardenability test.

This test is based on the principle that in hardening steel by quenching the more rapid the rate, the greater the heat extraction and the greater the hardness, while the slower the rate, the slower the heat extraction and the lower the hardness. The factors affecting the hardness attained are analysis and mass. For any given analysis, each rate of cooling will produce a hardness corresponding to the rate at which the heat is extracted.

In rounds of rather large size it has previously been shown that the hardness decreases from the surface to the center as a

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[1] *J.S.A.E.*, **50** (No. 1), 15–20 (1942). Tentative Method of End-quench Test for Hardenability of Steel, *A.S.T.M. Designation:* A255-42T.

[2] Republic Steel Corporation, National Emergency Steels. A pamphlet or booklet, by the corporation, not dated, but known to be printed in 1943.

result of a decreasing cooling rate. The effect of lower cooling rate at the center of large sections can be partially offset by the use of alloys. That this is possible is due to the fact that, with a given carbon content, alloying elements shift the *S*-curve to the right and thereby make it possible to produce high hardness even at the slow rate developed in the center of the large sections.

In interpreting the results of the end-quench test we make use of the previously discussed factors that basically indicate that cooling rate and hardness are one and the same. Cooling rate is the cause and hardness is the effect.

Since the 1-in. round face of the specimen was the only part quenched by the water, it is apparent that the rate of cooling will be rapid at the face and will then diminish gradually at distances away from it. The hardness will, as a result, gradually fall off. This decreasing hardness is the important factor in the test because somewhere along the length of the quenched specimen there are points that represent practically every quenching condition, from a water quench to an air cool and from surface to center of various shapes and sizes. This wide-range test, therefore, will show the degree of hardness that may be found on or in a wide variety of parts made from a wide range of analyses.

The correlation of this information is represented graphically in Fig. 22-X,[1] which shows the equivalent hardness at the center of various-sized rounds to the distance from the water-cooled end of the standard end-quench test bar. The plotted curve is merely a typical curve, but the other relationships—the equivalent hardness at the center of round bars quenched in still water and still oil to the distance from the quenched end of standard hardenability bar—have been established by Jominy.[2] The curve shows that the cooling rate at ⅜ in., ¾ in., 1$\frac{1}{16}$ in., and 1½ in. from the quenched end of the specimen results in hardnesses equivalent to those obtained at the centers of 1-in., 2-in., 3-in., and 4-in. rounds, respectively, when quenched in still oil.

The practical usefulness of the test is as follows: (1) to compare the hardenability characteristics of steel of difficult analyses, and (2) to produce hardenability data that can be used to predict

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[1] Republic Steel Corporation, National Emergency Steels.

[2] Interpretation of the End-quench Test, National Emergency Steels, A.S.M., 1942, pp. 30–32.

the hardness obtainable in any steel when that steel is to be used in machine parts not yet in production and not similar to any parts produced in the past.

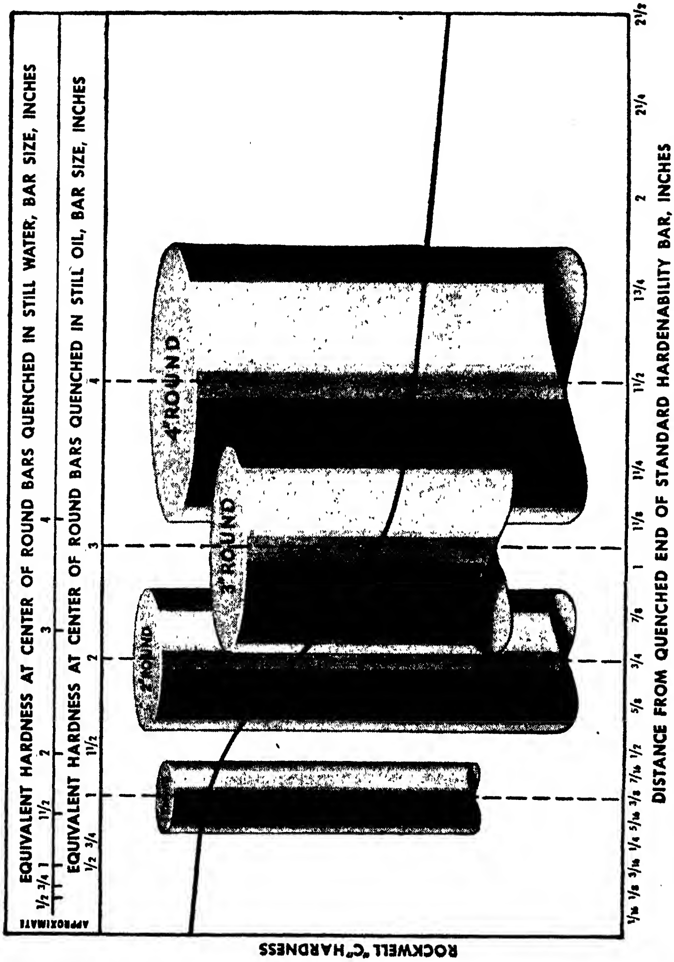

FIG. 22-X.—Equivalent hardness at center of rounds, end quenched hardenability test.

A comparison of the hardenabilities of four steels of different analyses is shown in Fig. 23-X. The curves clearly indicate that the depth of hardening is a function of the alloying elements, and that small quantities of several properly selected elements are more effective in influencing hardenability than a large quantity of a single element.

The curve for C1050 is typical of a shallow-hardening plain carbon steel in which the hardness values beyond about ¼ in. are low. The other curves show the varying hardening powers of small quantities of several properly selected alloying elements.

The curves also illustrate the previouly discussed principle that the maximum hardness is a function of the carbon content only. The maximum hardnesses for all the steels shown are for all practical purposes identical.

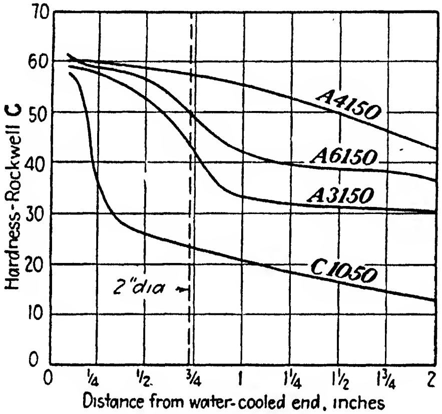

FIG. 23-X.—Hardenability curves (end quench bars) of pour steels containing 0.50 per cent carbon. Fine grained steel.

The correlation of the end-quench test with cooling rates at the center of a 2-in. round bar quenched in oil is shown by the vertical dotted line. The cooling rate at ¾ in. from the water-cooled end, as was previously shown, is the same cooling rate as that found at the center of a 2-in. round, quenched in oil.

The usefulness of the hardenability curves when one steel is substituted for another in a given part is illustrated in the following example:

A 2-in. round shaft made of A3150 steel was specified to be oil quenched and to show a minimum surface and center hardness of Rockwell C 50 and 40, respectively, as quenched. The shaft was quenched and found on sectioning to be 55 on the surface and 43 in the center. An end-quench test was made on the A3150 steel and C 55 fell at ⅜ in., while C 43 was at ¾ in. The subsequent procedure following the establishment of the key points was to select a steel having an end-quench curve that gave C 55 and 43 at 3/8 in. and 3/4 in., respectively. When such a steel was found, it would, therefore, be considered a satisfactory substitute on a hardenability basis for the A3150 steel.

Fracture-and-etch Test.—Another type of tests used to determine hardenability is the so-called *Shepherd P-F* (penetration-fracture) test.[1] This test consists in fracturing a hardened piece and observing the depth of hardening in relation to the unhardened core.

It is of interest to discuss briefly the reason for the possibility of showing the depth of hardening in a fracture test. When a bar of steel is hardened so that it will be martensitic on the outside and pearlitic in the center, a fracture of the piece will show a strong contrast between the two zones. This condition is due to the fact that martensite breaks with a brittle fracture, while the unhardened center breaks in a ductile manner. The transition from rim to core is quite sharp and corresponds to the position on the hardness distribution curve, where there is a sudden drop in hardness. Grossmann[2] has shown that this transition occurs at about the position where the steel consists of approximately half pearlite and half martensite and is in the region of most rapid formation of pearlite. It is this sudden change from predominantly pearlite that makes it possible to read the depth of hardening. Thus, the more pronounced the change in pearlite, the easier it is to read the depth of hardening from the fracture. With some steels and with some sizes, the gradient in hardness is quite gradual, so that some difficulty will be encountered in reading the depth by this method.

Microstructure.—Miscroscopic examination of the polished and etched specimen is another method for determining the hardening. This method is based on the fact that the hardness distribution is influenced in a regular manner by the proportion of pearlite present. It is evident, then, that the etching behavior of a polished cross section will follow this relationship. Thus, the more pronounced the change in proportion of pearlite, the

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[1] SHEPHERD, B. F., "The P-F Characteristic of Steel," Trans. A.S.M., Vol. 22, pp. 979, 1015, 1934.

[2] GROSSMANN, M. A., Hardenability, As It Affects Heat-treated Parts, *Metal Progress,* **33,** (No. 4), 373 (1938).

easier it will be to determine the depth of hardening by this method.

Factors Affecting Depth of Hardening.—In the carbon steels, the effects that manganese and silicon in the amounts usually present have on the hardening characteristics must be considered. The effect of such alloying elements is to decrease the critical cooling rate necessary to produce full hardening, thereby making it easier to produce the full hardness as determined by the carbon content and increase the depth of penetration in a given section size. The amount of alloy in the plain carbon steels will produce only a relatively small effect in this direction as compared to the large quantities present in some alloy steels. The effect of the various alloys on the hardenability will be discussed in detail in Chap. XVII.

The depth of hardening is affected also by the austenitic grain size of the steel. It has been shown that the fine-grained steels harden less deeply than the coarse-grained steels. This indicates, then, that fine-grained steels have a faster rate of cooling than the coarse-grained steels. This effect will be discussed in detail in Chap. XIII.

The structure of the material prior to hardening also has an important effect on the hardening. This effect is evidenced in the cases where the structure is nonuniform, as in banded structures, structures resulting from improper heat-treatment prior to quenching, etc. The results of such conditions will be nonuniform in both penetration and hardness.

The condition of the surface is obviously important, as the rate of heat removal is directly dependent upon the unimpeded flow from surface to quenching medium. Such things as tightly adhering scale or soot prevent the rapid transfer of heat to the quench bath, with the result that the critical cooling rate and full surface hardness will not be attained.

Suggested Questions for Study and Class Discussion

1. *a.* Draw a typical *S* curve and explain its significance.
   b. What laboratory procedure can be used to develop such a curve?
2. *a.* Make a progressive list of the structures that are produced in a eutectoid steel by cooling the steel from above the critical and allowing the transformation to take place at progressively lower temperatures, and those that are produced by tempering at progressively higher temperatures from the fully quenched condition.

- *b.* Would all these structures be obtained in practical quenching and tempering operations?

3. Define the terms:
- *a.* Pearlite
- *b.* Nodular pearlite
- *c.* Bainite
- *d.* Martensite
- *e.* Troostite
- *f.* Sorbite
- *g.* Spheroidized cementite
- *h.* Critical cooling rate

4. *a.* Would you quench a 0.50 per cent carbon steel from the same temperature as a 1.15 per cent carbon steel? Why?
- *b.* Would the hardness of the 1.15 per cent carbon steel be greater than the 0.50 per cent carbon steel in the as-quenched condition? Explain.
- *c.* What would be the general trend of the hardness and strength properties on tempering?

5. Explain the significance of annealing, normalizing, and quenching with respect to the *S* curve.

6. What is austempering?

7. What is meant by the term "hardenability"?

What is meant by a steel's having low hardenability? High hardenability?

8. By what methods are we able to determine the hardenability of a given piece of steel?

Describe the hardness-gradient and Jominy test methods for determining hardenability.

9. What effect does mass have on hardenability?

10. What factors affect the depth of hardening of a given steel?

CHAPTER XI

THE HEAT-TREATMENT OF PLAIN CARBON STEELS

The heat-treatment of ferrous alloys is defined in the "Metals Handbook" as "an operation or combination of operations, involving the heating and cooling of a metal or alloy in the solid state for the purpose of obtaining certain desirable conditions or properties." Heating and cooling for the purpose of mechanical working are excluded from this definition.

It must be noted, however, that cooling from hot-working operations, such as forging or rolling, may, and frequently does, constitute a heat-treatment just as truly as if the material had been subjected to a heating and cooling cycle. The structure and properties resulting from such cooling will for many purposes not be satisfactory, in which case an additional heat-treatment is necessary.

The desirable changes in conditions that are brought about by heat-treatment may be considered as being due to changes in the structure. These changes effected by heat-treatment involve changes in internal stresses and changes in the nature, form, size, or distribution of the structural constituents. Such changes in the nature of the constituents resulting from the effect of temperature on phase equilibria will be discussed in detail under the respective treatments.

HEATING AND COOLING CYCLES

Heat-treatment consists in subjecting a piece of metal to a definite time-temperature cycle involving two fundamental operations: a heating cycle and a cooling cycle. The time of holding at one or more temperatures in the heating and the cooling cycle is a third factor in the operations and must be considered in setting up a schedule of heat-treatment for any given metal or alloy.

Heating.—The first step in the heat-treatment of steel is the heating of the material to some given temperature in order to bring about the transformation desired. Frequently the term "heat-treatment" is considered, in a more restricted sense, simply to mean the heating and cooling operations involved in the hardening and tempering treatments. In this discussion we shall consider all heating and cooling operations as a heat-treatment, regardless of their purpose.

In all heat-treatments, with the exception of rare cases of differential hardening or local softening, satisfactory treatment involves uniformity in heating throughout to the proper temperature and uniform cooling at the proper rate. Absolute temperature uniformity throughout the entire heat-treating operation can never be achieved, owing to the fact that heat always flows when there is a temperature differential and will remain unchanged only at a given temperature, both inside and out, if its environment is at that same given temperature. It is essential, however, that there be either the addition or the removal of heat, so that any change in the environment temperature will bring about a heat flow to or from the piece.

Heat Flow.—In the heating of a piece of steel, the heat is imparted to the object by the processes of radiation and conduction (except where the piece is utilized as an electrical resistor) and convection which is an accompaniment to conduction. The actual flow of the heat, whether it be heating or cooling, takes place through the surface[1] and is dependent upon the rate of temperature change at the surface and the thermal conductivity and specific heat of the material. The outside surface of the piece of steel in heating is continuously absorbing heat from the furnace gases until it finally reaches the same temperature as the gases; then it stops. The center of the piece can get its heat in only one way, by conduction from the surface. It is this differential of surface and center temperatures that causes the heat flow. In order to have this heat flow, the center of the piece must always be colder than the surface until the entire mass reaches the furnace temperature. In addition, the surface temperature must also be below the furnace temperature; otherwise it would be impossible for the surface to absorb heat. The only way, therefore, that the entire piece can be uniformly heated is for the center to stop cooling the surface. Under such con-

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[1] The surface in this case refers to the solid surface of the heaviest section of the piece and does not refer to sharp corners, fragile projections, etc.

ditions, the center and the surface must arrive at the furnace temperature simultaneously.

Figure 1-XI shows a typical heating curve for a eutectoid carbon steel in which the solid curve follows the heating of the surface and the dotted line, the center. Do not scale this curve as it is shown merely to serve as an example of conditions that may hold true in any furnace.

There are several important points to be gathered from the trend of this curve. First, the general shape of the curve indicates that the heating is rapid at first and very slow at the end. This is due to the fact that at first the heat differential is greatest and that it gradually decreases as the furnace temperature is approached. The actual temperature difference according to the curve is approximately 100° during the early stages and is much smaller as the critical point is approached.

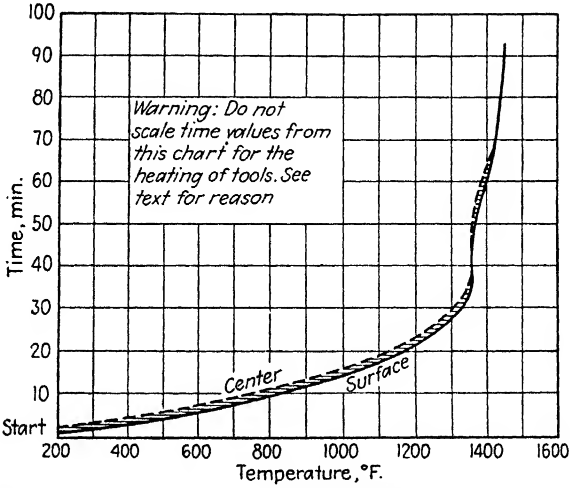

FIG. 1-XI.—Heating curve for a eutectoid carbon steel, 3 in. round × 6 in. long. Electric muffle furnace maintained at 1450°F. (*Courtesy of Carpenter Steel Company, Carpenter Service Bulletin* 6, 1937.)

The flat spot at 1360°F. marks the critical point of the steel. Since a considerable amount of heat is necessary to complete the transformation, the steel stops its temperature rise until this amount of heat has been absorbed.

The other point of interest is to note that the curves run together during the last 25 or 50°. It has been found that this trend will hold for all sizes and for all temperatures of heating. This, therefore, shows that when the surface is at the heating temperature, the center will also be at temperature.

It must be emphasized that the preceding discussion of heat flow is the ideal condition. Under actual operations a number of factors serve to complicate the problem, so that the ideal condition is probably never achieved.

First, the material may have a certain inherent resistance to heat flow, which we can do nothing about. Because of this difficulty, large sections will fail to harden completely and other sections will crack on heating, owing to the fact that when the heat is applied faster than it can be conducted through the piece there will be local thermal expansions or volume changes at the critical transformation temperature instead of uniform transformation throughout.

The shape of the section as well as the size affects heat flow. The sphere is the ideal section for uniform heat transmission, but even the sphere must be uniformly heated or cooled or it will not gain or lose its heat uniformly.

The arrangement of the charge in the furnace is another factor. If a number of pieces are charged in a furnace, the mutual influence of each section definitely affects the rate of absorption and loss of heat. The ideal arrangement of the charge in the furnace should be such that the heat source radiates its energy to the entire surface of every piece at the same rate. The same principle applies in cooling, in that each piece should be completely surrounded by the cooling medium so that each piece will lose its heat at the same rate.

The heating furnace, itself, is a large factor in the control of heat flow. It should be designed, as nearly as possible, to heat the entire charge to a uniform temperature. However, to ensure that all parts of the furnace shall be at the same temperature and also ensure that the charge will be heated to a uniform temperature presents a considerable problem.

Maximum Temperature.—The selection of the maximum temperature in the heat-treating schedule is determined by the chemical composition of the steel, the proportion of the phase (carbide or ferrite) that it is desired to take into solution into the austenite, and the specified austenitic grain size, if such a condition is considered. In general, the higher the temperature above the critical, the greater will be the solubility of the phases and the more rapid will be their rate of solubility into the austenite. In addition, the higher the temperature, other things being equal, the larger will be the austenitic grain size.

The general recommendation for maximum temperature in the austenitic range is that it shall be just high enough to bring about the desired amount of solution of the phases. In addition to the temperature, consideration must also be given to the amount of time at the optimum temperature in order that the austenite shall become essentially homogeneous and produce uniform results on subsequent cooling and also avoid, as far as possible, excessive austenitic grain growth. For hypoeutectoid carbon steels this means heating to a temperature just above the A_3 critical, which temperature allows complete solution of the carbide and proeutectoid ferrite with minimum growth of the austenitic grains. Any higher temperature will cause austenitic grain growth (except in grain-controlled steels); however, this may be desirable to improve machinability or to effect deeper hardening. These factors will be discussed in detail in Chap. XIII.

For hypereutectoid carbon steels the recommendations are to heat to a temperature between the lower critical (A_1) and the upper critical (A_{cm}). Since the critical rises so steeply with increase in carbon content, an excessively high temperature is necessary to bring about complete solution of the proeutectoid cementite, the result being the development of coarse austenitic grains with their accompanying effects of deep hardening and cracking. At temperatures just at the A_{cm} or below, but above the A_1, some of the carbide will remain undissolved, the amount varying with the temperature used. The undissolved carbide under these conditions restrains the austenite grain growth and acts as a nucleus for the transformation on cooling; such control of grain size, however, results in a sacrifice of the potential hardenability of the steel. The temperature to be used, therefore, is the result of a balance between the length of the heating period and the results desired.

The maximum temperature in a heat-treating schedule may also be determined by additional requirements, such as softening, producing a specified combination of strength and ductility, and the relief of internal stresses. The treatments involved in producing these requirements use temperatures that range from the recrystallization temperature of the material (recrystallization temperature for iron, approximately 840°F.) to just below the critical range (below A_1). A more detailed study will be made later in the text of the treatments involved, which are, briefly, spheroidization or low-temperature anneal, which consists in heating for a considerable time at temperatures below the A_1 critical to produce the desired ductility and softness in low carbon steels and easier machining in high carbon steels; low-temperature anneal, to bring about recovery (change in internal stress), recrystallization, and grain growth, which all lead to softening in cold-worked low carbon steels; and tempering, which is a low-temperature treatment of hardened steels used to relieve quenching stresses and to work for the recovery of a limited amount of toughness and ductility.

Measurement and Control of Temperature.—The precision with which we are now able to measure and control the temperatures developed in our modern heat-treating furnaces has been a vital factor in the development of accurate heat-treating methods on a large-scale production basis.

The problem of control with reference to quality involves control of the supply of fuel and air or electricity to the furnace, control of the supply of heat to the furnace, control of the furnace atmosphere, control of the manner of exposing the work while heating or cooling, and control of the rate and time of heating or cooling.

When we speak of control of temperature of a heat-treating furnace we should not make the assumption that this means control of the temperature of the heated product. In other words, there is a distinct difference between the control of heat as it is generated and the control of heat as it is applied or transferred from the product being heat-treated. A uniform temperature is essential, as we know, to the production of a uniformly heated or cooled product, but heating the furnace uniformly does not mean a uniformly heated or cooled charge, although it is the first step in the right direction.

Furnace-temperature control is brought about by the use of pyrometers in conjunction with certain methods of applying the fuel or electricity to the furnace. Such a control is useless unless we also consider the design of the furnace, *i.e.*, the effect of design with reference to uniform heat application to the charge.

In regard to pyrometer records we should always remember that this record is merely an indication of the temperature at that part of the furnace chamber in which the thermocouple is located and does not, necessarily, indicate the temperature of the product. The use of the pyrometer and other control devices is merely a means to an end. The real test or testimony is not the temperature-indicating chart but actual physical properties developed. As Bullens states, either the product is uniform or it is not. If it is not, then the testimony of the pyrometer chart is outweighed by that of the testing machine.

Heating Rate.—The general consideration regarding heating rates is that the rate shall be scheduled so as to prevent injury to the material through excessive thermal and transformational stresses. The condition of temperature uniformity and minimum damage to the steel resulting from stresses can best be obtained by slow heating.

The time required to heat a piece of steel is influenced by a number of factors, such as the initial condition of the material, the initial furnace temperature, the furnace atmosphere, the rate of circulation of the heating medium, and the size and shape of the section.

The initial condition of the steel with respect to the internal stresses produced by such operations as hardening or cold working has an important bearing on the most desirable heating rate, because of its effect on distortion and cracking and its effect on the initial size and growth of the austenite grains. Highly stressed, hard, and brittle materials, such as those produced by fully hardening, should be heated more slowly and carefully than materials that are stress free. At the other extreme, we have those materials that have been subjected to certain amounts of cold work that have a tendency toward excessive grain growth when heated very slowly through the transformation or recrystallization range. In such cases it is necessary to increase the heating rate materially when passing through this particular range, in order to avoid excessive grain growth.

The effect of initial furnace temperature on the time of heating should be obvious from our previous discussions of heat flow; the hotter the furnace, the shorter will be the total heating time. The exact time for a given furnace with a given charge must be determined by experiment. It is important to keep in mind one point regarding the initial temperature and that is, not to use a heating temperature at or within the critical transformation range when heating for subsequent hardening operations. This may be explained by citing an example in which two pieces of eutectoid carbon steel of the same size and analysis with a previously determined critical temperature of 1360°F., are heated in similar furnaces, which are set at temperatures of 1370° and 1450°F., respectively. Each piece will reach 1360°F. with but a small difference in total time, the one at 1450°F., however, taking the shorter time. At this point, however, a big difference will be found. Both the pieces will require a certain amount of heat to complete the transformation, but the one at 1450°F. will have a temperature differential of 90°F. while the one at 1370°F. will have only 10°F. It will be found that, in order to complete the transformation in both steels, it will take ten times, approximately, as long for the completion of the transformation in the 1370° furnace as in the 1450° one. The time required to absorb the heat necessary is dependent upon the rate at which the heat is being applied; the rate, in this case, is dependent upon the temperature differential. This is the reason why all commercial heat-treating recommendations specify heating to at least 25 to 50°F. above the critical transformation range.

In using furnaces that are not equipped with atmospheric control, the practice is generally to increase the heating rate above the ideal, in order to reduce the amount of scaling. This practice is also followed in order to secure greater production from a given furnace. If the atmosphere in the furnace is controlled, scaling can, of course, be minimized or eliminated entirely.

The rate of circulation of the heating medium over the surface of the steel has a definite effect not only on the time of heating but also on the uniformity of heating and the amount of fuel necessary to heat. In heating at temperatures of below approximately 1000°F., most of the heating is accomplished by conduction and convection and, in order to promote rapidity and uniformity, it is desirable to have forced circulation or convection of the furnace atmosphere. Forced convection can be secured by moving the work, suitably positioning the work, producing drafts through the furnace by suitably positioning the burners and exhaust ports, producing a circulatory action in the bath, as in salt-bath furnaces, and by the use of fans as in the case of muffle, radiant tube, or electric furnaces.

The reason for the predominance of convection heating at the lower temperatures can be found by considering several basic facts on heat transfer. Convection is basically dependent upon the temperature difference between the atmosphere and the charge being heated or cooled and on the rate at which the atmosphere flows past the surface. The rate of heating by convection, however, is practically independent not only of the temperature difference but also of the temperature of operations, owing to the fact that it varies as the difference in the fourth power of the absolute temperature. According to Trinks,[1] this means that "1°F. temperature difference at 2200°F. will cause heat to be transmitted 6½ times as fast by radiation as it will at 1200°F., whereas the heat transferred by convection by 1° difference of temperature is but little affected by the temperature." At high heating temperatures, then, radiation is the principal means of heat transfer, circulation also being necessary when the charge is so located that certain parts cannot be heated by radiation.

The size and shape of the piece of steel, too, will have its effect on the time required for heating. Regarding the shape of the section, it is rather obvious that it will take longer to heat a large section than a small section. By noting the rate at which different-sized rounds heated, it has been found[2] that the time required to heat a solid cylinder is directly proportional to its diameter. This rule merely gives a rate relationship for a given size and obviously it is necessary to experiment further to determine the rates for other shapes. Table 1-XI (according to Howard Scott) shows the relative heating rates for various shapes compared with a long cylinder. These factors have been determined by mathematical calculation, but it has been found by actual test that the values check quite closely.

Application of the information in Table 1-XI is made by first determining the length of time necessary to heat a cylinder under the conditions desired and then determining the time for other shapes by dividing the time for the cylinder by the "speed factor"

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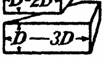
[1] Trinks, W., "Industrial Furnaces," 3d ed., Vol. I, p. 382, John Wiley & Sons, Inc., New York, 1934.

[2] *Carpenter Service Bulletin* 8, Carpenter Steel Co., Reading, Pa.

for the new shape. For example, if a long cylinder 1 in. round takes 30 min. to heat, a long flat 1 by 3 in. would take 30 ÷ 0.6 or 50 min. to heat.

The shapes and sizes just discussed have been relatively uniform, but in many cases there is considerable variation in the dimensions. In practical heat-treating operations, the problem of variable temperature rise within thick and thin sections of articles of variable cross section must be recognized and, whenever possible, precaution should be taken to retard the heating in the thinner sections so as to keep distortion and thermal stress at a minimum. In addition, proper precaution must also be taken to prevent, as much as possible, the corners of the section from reaching the final temperature for too long a time before the major part of the mass.

TABLE 1-XI.—RELATIVE HEATING SPEED OF VARIOUS SHAPES COMPARED WITH A LONG CYLINDER

	Shape	*Speed factor*
	*Long cylinder (dia. = D)*	*1.0*
	*Long square (D x D)*	*1.0*
	*Long rectangle (D x 2D)*	*0.7*
	*Long rectangle (D x 3D)*	*0.6*
	*Infinite plate (very wide, thickness = D)*	*0.5*
	*Sphere (dia. = D)*	*1.5*
	*Cube (D x D x D)*	*1.5*

Holding at Heat.—The general requirements of time at a given temperature are that the time shall be long enough to attain the proper temperature uniformity and bring about the constitutional changes that are desired. Time at a given temperature must, however, be kept at a minimum consistent with several requirements.

In all the heat-treating operations, temperature and time must be considered simultaneously, since the influences of the two are similar. From a temperature viewpoint, practically all changes in metals take place more rapidly the higher the temperature, because of the increased atomic mobility, while with increased length of time at a given temperature the greater will be the amount of elements taken into solution and the more homogeneous the material, because of the diffusion of the elements.

The ultimate grain size of the material is also governed by the time-temperature relationship. At a given temperature there apparently is reached an equilibrium grain size that changes very slightly with an increase in time. The time necessary to reach this particular condition is dependent upon the temperature and ranges from an extremely short period near the melting temperature to a matter of days just above the so-called recrystallization temperature.

In general, as far as the desired constitutional changes are concerned, the element of temperature is more important than that of time. This is due to the fact that a small increase in temperature will have a much greater effect in bringing about the desired changes than a longer time at a lower temperature.

The time factor, however, becomes considerably more important when other requirements are considered. First, the time must be sufficient to permit the entire section to attain a uniform temperature since, in the majority of cases, uniformity is highly essential. Time to attain this temperature uniformity increases with the thickness of the heaviest section of the piece or the lower the thermal conductivity of the material. There is also the effect of scaling and decarburization—reactions that take place continuously with the length of time at temperature. Many of these reactions can be controlled by the use of proper atmospheres or utilized advantageously as in carburizing, nitriding, etc.; but if the effects are in any way harmful it becomes necessary to keep the time as short as possible.

Cooling.—Cooling, in the iron-carbon alloys, is a quite important factor, particularly in the rates of cooling at each stage in the heat-treating operation. It is essential to remember that the principles governing the transfer of heat from the material in cooling are the same as those governing the transfer of heat to the piece. The essential difference is the rate of transfer, which is dictated by the structural requirements made of the final cooled piece or by the physical characteristics, such as the size, shape, and surface condition of the piece to be cooled. We might say that the rate of cooling is more important than the rate of heating.

DESIGN

It is not difficult, after the previous discussion, to realize the importance of proper design. Take, for example, the two sections shown in Fig. 2-XI. The *L*-shaped section is quite uniform in cross section, but it is found that the heating and cooling will not be uniform. Investigation has shown that those parts that cool most rapidly will usually heat most rapidly. For example, the protruding corner *A* will, in cooling, give off its heat from three sides, while it has been heated from one side only. This corner will, therefore, cool faster than an edge, such as *B*, which is cooled on two sides and has been heated on two others. A point on the flat surface, as *C*, receives heat on one side and delivers it from the other. Point *D* is exposed on one side only to the cooling action but is surrounded on three sides by hot metal and will naturally cool last and most slowly. Thus, in such a section differential cooling will result and heavy strains will be set up at the sharp angle at *D*, which is last to cool and will most likely crack. If the cracking does not occur during the cooling, it is almost certain to take place in service, which will concentrate additional stress at this point. That danger may be lessened if the angle at *D* is replaced by a rounded fillet, which will prevent the concentration of stresses at the sharp angle.

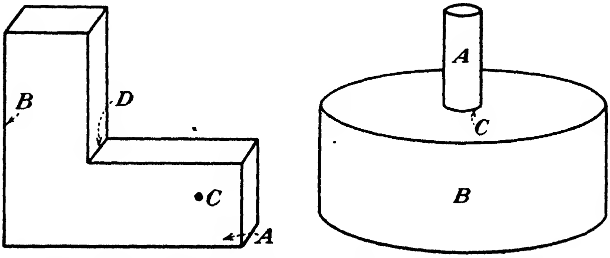


FIG. 2-XI.—Effect of design on the rates of cooling.

If, in addition to having sharp angles, the object has one part massive and another thin in comparison, as in Fig. 2-XI, the difficulty is further increased. The light part *A* will, of course, cool faster than the massive part *B* and it will be almost impossible to prevent cracking at the angle *C* under very rapid rates of cooling. Similar difficulties will be experienced if heating has been too rapid. In many cases sharp angles, as at the root of a gear tooth, a cutting tool, or a keyway, must be rounded off into a fillet. Parts that need not, otherwise, be heavy must sometimes be made heavy on this account, or massive parts must be made hollow to compensate for adjoining thin parts, so that cracking will not occur. For more detailed information on the relation of design to heat-treatment see the reference in the footnote.[1]

QUENCHING PRACTICE

In all the previous studies of the heat treatment of steels there has been, as the basic principle, the production or the inhibition of equilibrium. The engineering principles thus involved were the heating and cooling operations, which included a study of the rate of flow of heat from a hot metallic object to a bath during quenching or from a furnace into a cold metallic object during heating. The extent to which certain phase changes take place is dependent on the velocity of the phase change in the temperature regions passed through in cooling and on the velocity of heating or cooling of the object.

The rate of cooling to bring about the desired properties in annealing should be well in mind, so let us now direct our attention to those media that produce a rapid rate of cooling during the quenching operation in order satisfactorily to harden the steel at some depth below the surface.

The rate of heat transfer from a section of steel that is being quenched from the usual temperature range into a medium at or near room temperature is dependent upon the following factors: the size and shape (weight of the piece, its thickness, the amount of metal per unit of surface, etc.); the thermal or heat conductivity of the metal; the specific heat of the metal; the surface condition; and the properties of the quenching medium used. The fundamental properties of cooling liquids that dominate their cooling characteristics are the boiling point, heat of vaporization, viscosity, thermal conductivity, and specific heat.

Quenching Stages.—For the sake of simplifying, let us consider what happens when a steel object that has been heated to a temperature of 1472°F. (800°C.) is plunged into a stationary bath

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[1] A.S.M. "Metals Handbook," p. 214, Cleveland, Ohio, 1939.

of cold water. The quenching action may be considered as taking place in three stages.[1]

First Stage.—In the first stage after the start of the quench, which is characterized by *A* in Fig. 3-XI, a continuous vapor blanket is formed, which envelops the entire specimen. In this stage no liquid touches the metal surface, the heat escaping from the surface very slowly by radiation and conduction through the layer of water vapor to the liquid-vapor interface. This is an undesirable condition in most quenching operations. The water is unable to come into contact with the metal surface because of the insulating effect of the film of vapor. This is the familiar spheroidal state encountered when a drop of water falls on a hot stove; the water does not wet the stove because of the formation of the insulating vapor film. The heat extraction during this stage in the cooling is extremely slow, owing to the fact that the cooling is by conduction and radiation through the gaseous film.

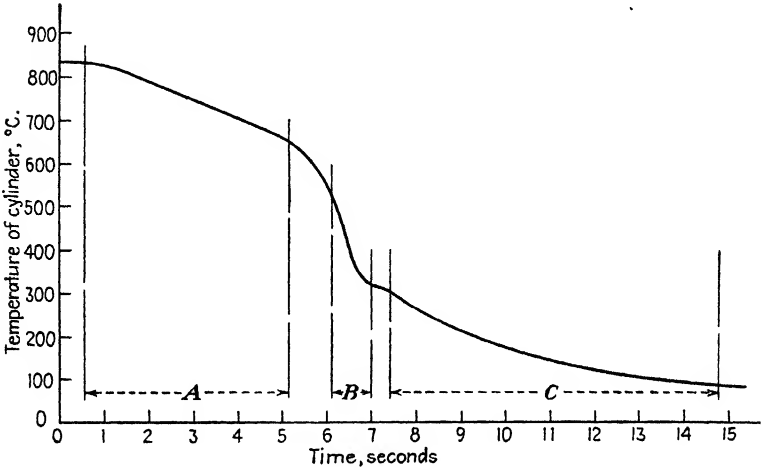

FIG. 3-XI.—Typical cooling curve for the center of a small cylinder during quenching in a still liquid which shows three stages of surface cooling. (*After H. Scott.*)

Second Stage.—When the temperature of the surface reaches a point where the radiation and conduction are insufficient to maintain a vapor blanket, the second stage of cooling starts. This stage is chiefly one of vapor transportation, *i.e.*, the vapor

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[1] SCOTT, HOWARD, The Problem of Quenching Media for the Hardening of Steel, *Trans. A.S.M.*, **22**, 577–604 (1934).

blanket is broken intermittently, allowing the liquid to wet the metal surface at one instant; then at the next instant, it is pushed back by forming vapor bubbles. The bubbles are carried away by gravity and convection currents and the liquid again touches the surface. This period, which is marked *B*, is a very rapid one, as is indicated by the steep slope of the cooling curve. This rapid cooling soon brings the temperature of the surface below the boiling point of the quenching liquid, vaporization can then no longer take place, and the third stage of cooling by convection begins. The rate of cooling in the second stage makes this the most active stage in the transfer to heat, owing to the great quantity that can be carried away by the vapors of many liquids, particularly those of water.

Third Stage.—The third stage is one of cooling by simple convection and conduction by a liquid in contact with a metallic surface whose temperature is below the boiling point of the liquid.

Since, in quenching, the maximum cooling rate must be developed very shortly after the start of the quench, the duration of the vapor-blanket stage is very important. The most important factor affecting the duration of the first stage is the temperature of the quenching medium. The temperature, when dealing with water solutions, is of considerable concern in the first stage, as the cooling is slow and if it persists for more than a few seconds will allow time for the transformation to take place at the rapid transformation zone on the *S*-curve and will produce soft spots. It is highly desirable, therefore, to suppress the vapor stage as much as is possible, in order to avoid this difficulty.

The duration of the first stage in water solutions may be doubled or even tripled as the bath temperature is raised to near its boiling point. This is due to the smaller heat flow necessary to maintain the vapor blanket, the nearer the liquid temperature approaches its boiling point. This effect is most prominent in liquids with low and sharp boiling points, such as water.

To counteract the persistence of this vapor stage, several measures[1] may be taken such as the use of a cold medium to shorten the life of the vapor and agitation, such as a spray. The most commonly used procedure, however, is to add either salt, NaCl, or caustic soda, NaOH. These additions, when added in

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[1] "Practical Metallurgy for Engineers," E. F. Houghton and Co., Philadelphia, 1943.

optimum amounts—9 per cent salt or 3 per cent caustic soda—apparently are effective through the release of their vapors from the solution during the quench. Upon release at the hot-work surface, the bursting and crackling of the salt crystals destroy the stability of the vapor film and thereby promote rapid early cooling. For maximum efficiency, the concentrations should be used as given above, because increasing the amounts lowers the cooling rate in the second stage. Thus, it is evident that a brine bath is not inherently a more severe quench than plain water, but that it merely permits the full use of the latent cooling power of the water by removing the interfering vapor phase.

The addition of soap to water has an adverse effect, in that it considerably increases the duration of the vapor phase. This is probably due to the stabilization of the water-vapor interface, which retards the release of the vapor from the metal surface.

In view of these factors, several rules of quenching should be observed:[1]

1. Keep water-base quenching liquids cold, preferably under 70°F.
2. Avoid contamination of aqueous quenching liquids with soap.

The vapor stage of cooling is found to be less important when oil is the quenching medium than it is when water is used. Actually, this stage may be longer, but this is not harmful unless it is exceptionally long, because the rate of cooling is faster in this stage with oil than with water. The duration of the stage is longer the lower the boiling point of the oil, but, in general, oils used for quenching have such high boiling points that the duration of this stage is little affected by an increase in bath temperature. Oils, in contrast to water solutions, can be used warm without danger of incomplete hardening and with advantageous effects, as will be noted subsequently.

The importance of the rate of cooling in the second stage is emphasized if it is noted that it corresponds to the temperature range of 1112 to 932°F. (600 to 500°C.) in which the steel in the austenitic condition transforms most rapidly, as is shown in the diagram in Fig. 7-X.

The cooling rates of various media taken at this temperature range will, therefore, indicate their relative cooling power in this

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[1] **Scott, Howard, Quenching Media, "A.S.M. Metals Handbook," p. 328, Cleveland, Ohio, 1939.**

stage. Water and brine solutions have nearly equal cooling power and it is the highest cooling power in this stage, while oils are much lower and differ considerably among themselves. Bath temperature does not materially affect the rate of cooling when a water solution is used. The cooling power of warm oil is higher than that of cold oil. This is probably due to the lowered viscosity, which permits more active movement of the vapor that is carrying the heat from the metal surface. The heat of vaporization of the quenching medium is the dominant factor in this stage.

In the third stage, where cooling is by convection and is controlled by the thermal conductivity of the medium, the differences between relative cooling rates of water and oil are greater than in the second stage. The high cooling power of the water solution is not essential, but it is tolerated because of its exceptional power in the second stage, where a fast rate is essential for successful hardening. In some types of steels, however, there is danger of cracking by rapid cooling into the zone where the hardening transformation takes place.

This danger of cracking is due to the fact that in steels containing more than about 0.50 per cent carbon the martensite formed on quenching is extremely hard and brittle and is very highly stressed. The stressed condition is due principally to the volume changes that accompany the gamma to alpha transformation. In this condition the steel is ready to relieve itself of these stresses at the first opportunity; in fact, in some cases the stresses in fully hardened steel are of such magnitude that spontaneous cracking will occur. However, if the cooling rate is slowed down in this stage, the danger of cracking can be materially reduced.

Oils are particularly advantageous in this respect, as their cooling rates are materially slower in this stage than are those of water. The important advantage of the use of oil, however, is that the bath temperature can be raised without bad effects on its characteristics in the other two stages, but it will materially lower the cooling rate when the temperature of transformation to martensite is approached.

In view of these factors the following rule is suggested:[1] Keep quenching oil warm, preferably between 90 and 140°F.

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[1] Scott, Howard, Quenching Media, "A.S.M. Metals Handbook," p. 333, Cleveland, Ohio, 1939.

Factors in the Selection of a Quenching Medium.—There are many factors involved in the selection of quenching media, such as cost, stability of quenching characteristics, ease of removal from the work, corroding action, effect of scale adherence, fire hazards, fumes, and odor; but for lack of space we will concentrate our discussion on the fundamental properties that determine depth of hardening, distortion, and residual stress.

Depth of Hardening.—In some classes of work, high strength or hardness is required only to a moderate depth below the surface. Carbon steels serve well in these applications, because they cannot be hardened to any great depth, not over ½ in. in large sections, no matter how active the surface cooling may be. For full hardening of sections of over ½ in. round, an alloy addition must be made to the steel. The alloy content has the effect of decreasing the critical cooling rate and thereby permits slower cooling on increased section size, to produce full hardening. If interest were solely on the penetration of hardening, it would be possible to use the most active quench attainable and hold the amount of alloy addition to a minimum.

Distortion.—Distortion enters the picture in complicated shapes and can be controlled only by moderating the quench. Thus, in the selection of a quenching medium for such shapes, where full hardening is necessary, a compromise must be made between the antagonistic requirements of fast cooling with minimum alloy content and maximum machinability, and slow cooling with a higher alloy content for minimum distortion and lowered machinability.

Residual Stress.—Another important factor in quenching is the residual stress. This stress may be of such magnitude that cracks may develop, particularly when the conditions are such that tensional residual stresses are produced at the surface. This danger becomes greater with increasing carbon content. The conditions for cracking are most favorable when the temperature gradient is particularly sharp in the third stage of cooling or, as by comparison of temperatures, in the hardening transformation range, which in commercial steels varies between 400 and 200°C. (750 and 390°F.). Hence, in order that the possibility of distortion and cracking may be minimized without loss in penetration during hardening, the medium selected should be one that cools more slowly below 400°C. (752°F.), and as fast as possible at higher temperatures.

Hardening Practice.—In general steel hardening practice, two distinct types of quenching media are required, one for parts of homogeneous composition which require only superficial hardening, and another for parts which must be hardened fully without distortion. When superficial hardening is adequate and distortion is not a serious problem, plain carbon steels are used because of their low cost and good machinability. Except in extremely thin sections, an extremely active quench is required for carbon steels.

From what has been learned of the first stage of quenching, it is apparent that vigorous agitation of the quenching bath would be an effective way to prevent the formation of the vapor film. For example, if a high-pressure water spray is substituted for a still-water bath, the duration of the first stage is materially shortened or even prevented and the second stage is accelerated. The spray breaks up the vapor bubble formation quite rapidly.

Cold water with vigorous agitation is the medium most frequently used in cases where a very rapid quench is desirable. Direct agitation is not generally applicable in other than symmetrical shapes. For unsymmetrical shapes of otherwise favorable contour, a liquid is needed which, even though motionless, will produce surface cooling rates approaching those of a water spray. Common salt, as well as inorganic compounds dissolved in water, are favorable. Sulfuric acid is prominent in this regard and it also cools slowly in the third stage but is very hazardous to the workmen.

In complicated shapes where complete hardening is essential or when distortion is a major problem, an alloy steel is used. In these cases, the alloy has the effect of lowering the rate of transformation so that the medium used will not be required to cool as fast fully to harden a thick object that would not be fully hardened with any medium if it were made of a plain carbon steel. Oils—mineral, vegetable, or animal—are generally satisfactory for these sections and grades of steel because they provide reasonably fast cooling in the 600 to 500°C. (1112 to 932°F.) range and slower cooling at lower temperatures, which is a favorable characteristic, for it reduces distortion and the danger of cracking.

In order that the limitations of a single quench may be overcome, a double quench has been practiced, but the results are not generally reproducible or satisfactory. This duplex practice consists in the use of two liquids, immersing for a few seconds in one liquid, water, and following this by a transfer to another liquid, oil, for final cooling. It is preferable, however, to obtain a single liquid that would give the advantageous features of a double quench, namely, rapid cooling at high temperatures with slower characteristics in the low-temperature range.

General Characteristics of Quenching Media.—The major requirement of any coolant is that it should provide the desired rate of cooling. Its adaptability to commercial hardening practice is dependent upon other factors also, including low cost, general availability, permanence with respect to its cooling characteristics in continued use, maintenance of cooling characteristics with moderate temperature changes, ease of handling, and safety as regards workmen and handling. Failure to meet one or more of these requirements does not preclude its commercial use but it will influence the final selection. Failure to provide the desired cooling speeds automatically removes the coolant from further consideration.

In commercial heat-treating practice there are five coolants that are generally employed. These are (1) water in baths and sprays, (2) dilute sodium hydroxide solution, (3) sodium or calcium chloride brines, (4) oils or oil-water emulsions, and (5) air. These coolants will provide a wide range of cooling rates and, when used at ordinary temperatures with mild circulation (except as noted), will give speeds that will increase in the following order:[1]

Slow ↑ Air–motionless
Air–in motion
Oils
Water baths
Brines
↓ Dilute caustic soda and
Fast Pressure water sprays

TEMPERING PRACTICE

The change in the structure on heating a hardened steel, or the degree of tempering is dependent on the time and the tem-

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[1] "Nickel Alloy Steels," Sec. VI, Data Sheet No. 4, p. 5, International Nickel Co., 1934.

perature at which the steel is heated. This change may be roughly told by the color of the film of oxide that forms on a polished surface when the steel is heated in an oxidizing atmosphere. A very slight tempering may take place at room temperatures, but no surface discoloration will indicate this change. However, when a temperature of about 450°F. is reached, the iron oxide thus formed will give the metal a straw color. With the appearance of this color a decided tempering has taken place. Further increases in temperature of heating changes this oxide color to brown, purple, blue, etc.

The carbon content of the steel and the different alloys used in alloy steel change these colors and their temperature of occurrence. Since the advent of the study of metal structures by use of the microscope and reliable pyrometric temperature control of furnaces, this color method of judging drawing temperatures is of little use.

Many types of baths are used for tempering, as these ensure that the steel may be heated to a more accurate and uniform temperature than that produced by oven furnaces. Because of their cheapness and ease of handling, oil, salt, and lead baths are used more than any others. Parts that are to be tempered below 600°F. can be heated in oil, but above that temperature the oil will blaze and burn, thus rendering it useless for tempering. A further difficulty in the use of oil is that it must be removed if the tempered parts are to have a clean appearance.

Among the many salts used are barium chloride, sodium chloride, potassium or sodium cyanide, etc. The entire tempering range will not be covered by any one salt, but different ones are suitable in different temperature ranges.

Lead baths also are commonly used above temperatures of 619°F., the melting point of lead. Enough tin may be added to the lead to lower the melting temperature to nearly 300°F., thus giving a much wider range. Molten lead with proper precaution will not stick to the steel being heated, thus eliminating the necessity for cleaning.

The most recent development for heat-treating operations within the temperature range of 300 to 1300°F. is the forced-convection furnace. In these installations high-temperature fans are installed, either directly in the furnace chamber or in external chambers, and these serve to recirculate air or protective gas through the charge to accelerate heating and to ensure uniform temperatures throughout.

HEAT-TREATING FURNACES

The following discussion of heat-treating furnaces must be generalized because of lack of the space that would be required for studying the multiplicity of furnace designs available, even when only such factors as fuels, method of handling the work, and the nature of the work to be accomplished are considered. Furnaces are classified here according to the methods used in handling the material in its passage through the furnace. The two principal types are (1) the in-and-out, or batch, type and (2) the continuous type. In both, many differences will exist on account of the nature of the fuel used, whether it be coal, gas, oil, or electricity. The general principles, however, regarding the means of temperature control will in the main apply to all types, whether they be simple or complex. In general, it might be said of both types that the trend of present-day design is to make the furnace as simple and as automatic as possible, so as to eliminate the human factor as far as may be done. Furnaces must also be as economical as possible and, wherever it is practical, the continuous type of furnace should be used.

A question frequently asked regarding furnaces is, "Which kind is the best—a gas furnace, an oil furnace, an electric furnace, or a liquid bath?" In answering this question it should first be recognized that fuel used as the source of energy can be produced, on a B.t.u. basis, at a lower cost than can electrical energy. However, there is a considerable difference between the price of the fuel or electricity on a B.t.u. basis and the cost of heating. The cost of heating is the sum of the price of the heating medium and the amount required for the cycle. The amount of energy required is primarily a function of the thermal and mechanical characteristics of the equipment and its method of operation, regardless of the type of energy used. Hence, a comparison on a B.t.u. basis of fuel and electricity for heat-treating operations would be unfair unless it were considered along with the relative efficiencies of the different forms of heating equipment that may be adapted to the different forms of fuel or electricity and the specific job requirements. However, such rapid improvement has been made in the performance of all types of furnaces that, when this is taken together with the fact that different shapes actually require different types of heating, specific comparisons cannot be drawn.

The In-and-out or the Batch Type of Furnace.—This type of furnace, which is designated by such names as "box type," "batch type," and "in-and-out," is simply an oven with a door across the front. The stock to be heated is laid in a certain position on the hearth of the oven and remains there until it is heated. It is then removed, usually through the door by which it has entered. The size of this type of furnace varies with the size and quantity of work to be heated at one time.

In order to accomplish an even distribution of temperature throughout the furnace, a large number of burners are generally used, but even so it is practically impossible to heat steel uniformly, because in every case the heat transfer to the steel is made by radiation from the flame; hence, it is quite obvious that the side nearest the flame will be heated sooner and to a higher temperature than the opposite side. Recent types of heating that attempt to bring about a more even distribution of heat are the so-called *diffusion-combustion*, *surface-combustion*, and *electric-resistance* furnaces. In these there are large areas of incandescent bodies that distribute the heat quite evenly.

Probably the oldest furnace of this type is the coal- or coke-fired oven furnace. This consists essentially of two long chambers, which are parallel and are separated by a bridge wall that does not extend quite to the roof arch. One of the chambers serves as a firebox and has grates extending nearly its entire length; the other is the heat-treating chamber, proper. The flame direction can be seen by the direction of the arrow sweeping from the firebox, over the bridge wall, through the heating chamber and down through vents on either side of the chamber floor into the main flue and out the chimney. This old type of furnace presented a great deal of trouble with respect to uniformity of temperature throughout the chamber and, hence, the uniformity at which the work was heated.

Other batch- or box-type furnaces have been built in which the design of the furnace has been changed in an effort to get more uniform circulation of the hot gases. This condition should hardly be limited to the box-type furnaces, as this is and has been one of the major aims in the development of all the modern furnaces.

Figure 4-XI shows a side-fired batch furnace, *i.e.*, side-fired with respect to the position of the burners, the flame and the hot gases passing through the oven and down through vents on either side of the oven and under the hearth; an under-fired batch furnace, where the burners shoot beneath the hearth, the flames and gases passing up the sides and out through vents in the roof; an over-fired batch furnace, where the burners point into a combustion chamber above the oven of the furnace, the hot gases passing down through the oven of the furnace and out through vents on the sides; and, a direct-fired furnace, in which the burners point directly into the oven, with vents and flues at the bottom so as to cause the hot gases to pass through the oven.

The car-bottom type of furnace is merely a large "batch-type" furnace in which the hearth, or bottom, of the furnace is the floor of a flat car, which runs in and out of the furnace on a pair of rails. This type of furnace is particularly adaptable where heavy work is being heat-treated, on account of the convenience it offers in loading and unloading.

In all these types of furnaces, the products of combustion come into direct contact with the stock being heated. In cases where this contact is injurious to the stock, a muffle-type furnace is used, the stock in this type being placed in the muffle, which is heated by the products of combustion.

Furnaces that are heated by electrical means and that come under this particular furnace grouping have the characteristics of a muffle-type furnace. This classification is due to the constructional characteristics, in that the entire furnace is usually enclosed in a tight steel shell, with the doors and other openings arranged to close tight. In this way the atmosphere in the hearth is controllable and independent of any combustion problems.

The heat is liberated by means of resistor elements located on any or all sides of the charge in the walls, top, and bottom of the furnace. The resistors may be either exposed or covered with a refractory material that is a good conductor of heat but a nonconductor of electricity. The purpose of such covering is to protect the resistors from damage and short circuit with the

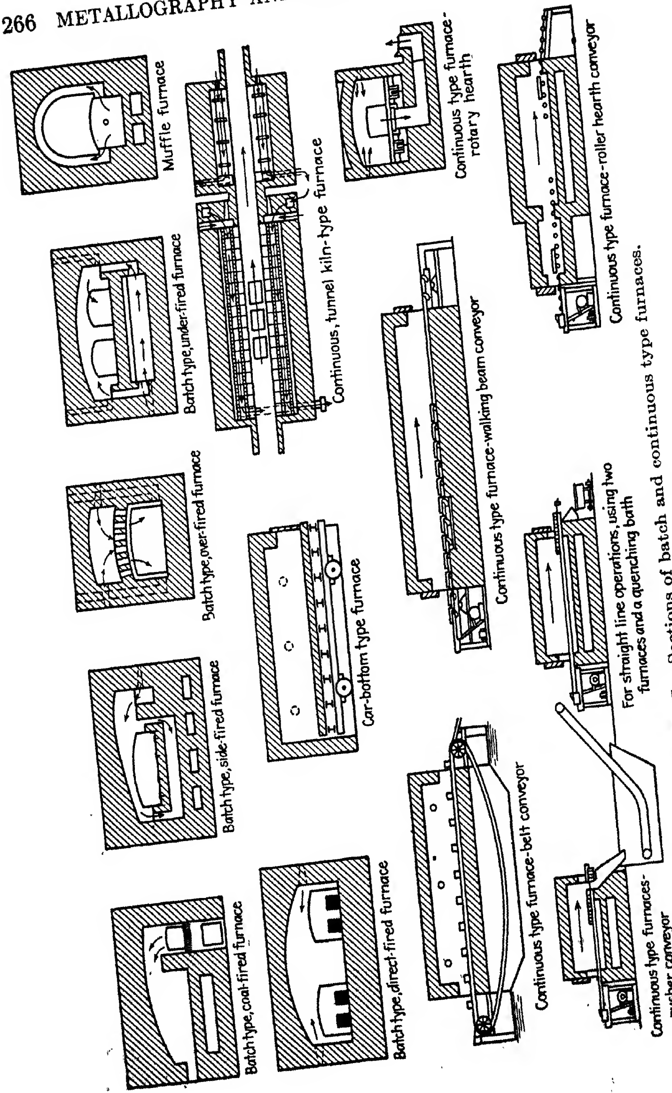

FIG. 4-XI.—Sections of batch and continuous type furnaces.

charge. The material in general use for resistors is an alloy of nickel and chromium of approximately 80 per cent nickel and 20 per cent chromium. This alloy, when heated in air, forms a resistant oxide coating that prevents further oxidation, and hence provides excellent resistance to deterioration within the recommended operating-temperature range. The maximum operating temperature of the heating chamber is usually specified to be 1850°F. For higher temperatures, up to 2400°F., nonmetallic resistors, such as carbon or silicon carbide, are employed.

Continuous-type Furnaces.—There are a number of kinds of this type of furnace, all of which are merely variations of a single principle. The cold stock is charged through a door at one end of the furnace and is advanced more or less automatically through the furnace to the discharge door. The problem of heating the cold stock is somewhat easier in this type because all the stock passes along the same path. This type of furnace is also very economical from a heat standpoint, because the hot gases from the hottest zone preheat the cold stock as it progresses from the charging door toward the heating zone. The furnace is generally divided into zones of preheating, heating, soaking, and —depending on the heat-treatment being effected—a cooling zone. The amount of heat or the temperature that is brought about by the use of some form of fuel or electricity is controlled by a set of thermocouples in each zone.

The method of conveying the material through the furnace is dependent upon the type of material that is to be handled. The car-bottom type is readily adaptable to a continuous-type furnace where there is sufficient production of work, all demanding the same type of treatment. Figure 4-XI shows a "tunnel" furnace single stream of the car-bottom type in which the work is charged through a door, advanced through the preheat, heating, soaking and cooling zones at a given predetermined rate of speed and is discharged at the door on the opposite side of the furnace; and other types of continuous furnaces using belt conveyors, roller hearths, walking beams, and the pusher type, which may be considered a general type for all of the above with the exception of the belt type.

In some types of heat-treating, two continuous furnaces are placed in a straight line with a quenching bath between the two. The cold stock is heated to its maximum temperature in the first furnace, dropped from that temperature into a quenching bath, carried by a suitable conveyor to the second furnace, where it is reheated or "drawn," possibly cooled, and then discharged. The first furnace and the quenching bath may in some cases constitute a separate heat-treating unit.

Another type of continuous furnace is the rotary hearth, or turret, type. In this case the cold stock is charged onto the hearth, which, as it rotates, brings the stock into zones of higher and higher temperature. The stock makes a complete circle through the furnace and is discharged near the charging door. This type of furnace is used for work that is to be removed from the furnace without any cooling from the highest temperature.

LIQUID HEATING BATHS

Liquid heating baths find their application in tempering, annealing, and hardening operations on small articles, such as drills, taps, chisels, dies, and parts of intricate shape. In almost all cases, liquid heating media now in use are either molten lead or molten salts. The molten salts or metal are contained in a metal pot, which is suspended in a refractory-lined oven; the heat is applied externally to the pot by oil, gas, or electricity. After the bath is heated to the desired temperature, the article to be treated is immersed for the desired time.

Molten baths have, in general, the following advantages:

1. More rapid transfer of heat into the charge at both high and low temperatures than in an oven type of furnace.
2. A very uniform temperature within the bath when, through proper design and operation, there exists good circulation of the heating medium.
3. Protection of the charge from the effects of the atmosphere.

Lead Baths.—The most common metal bath is lead, which has a useful temperature range between about 750 and 1600°F. The lower specified range is somewhat above the melting point of lead (620°F.) but some degree of temperature above this is necessary to prevent solidification of the lead on the work. The maximum temperature is limited by volatilization and pot life. It should be noted that lead vapors are poisonous, so that such a pot should be provided with a ventilating hood.

Since hot lead tends to oxidize when in contact with air, to produce an undesirable dross, 1 or 2 in. of granulated wood charcoal or a layer of molten proprietary salt is used on top of the bath to reduce this difficulty. Dross is undesirable for several reasons. It sticks to the sections as they are being immersed and continues to adhere to them as they are withdrawn and quenched. It also interferes with the quenching, is difficult to remove, and tends to decarburize the surface of the material. It should be noted, however, that clean lead has no chemical action upon steel, and will not oxidize, carburize, or decarburize it.

Salt Baths.—Liquid salt baths may be divided into two classes according to the type of salt used. These two classes are neutral salts and active salts.

Neutral Salt Baths.—Neutral salt baths are those intended to act simply as heating media and to have no chemical influence on the charge. There are available a large number of salts for such a purpose, covering temperature ranges from as low as 300°F. to as high as 2350°F. It must be noted that this temperature range covers a number of salts or combination of salts, as no one salt is suitable over the entire range.

For very low temperatures a combination of sodium nitrite, $NaNO_2$, and sodium and potassium nitrates, $NaNO_3$-KNO_3, is satisfactory, while for temperatures up to 1000°F. the sodium and potassium nitrate combination can be used. For temperatures of from 1100 to 1650°F., the commonly employed salt baths are combinations of sodium carbonate, Na_2CO_3, sodium chloride, $NaCl$, and calcium or barium chloride, $CaCl_2$-$BaCl_2$. For very high-temperature work, calcium and barium chloride mixtures or straight barium chloride may be used.

Several precautions must be observed in using the above salts. The nitrate salts are strongly oxidizing when used at temperatures above 1000°F. The commercial chlorides normally carry a small amount of sulphate, which is corrosive and must be eliminated. The chloride and chloride-carbonate mixtures have some tendency to corrode and to pit steel, and they also have a strong decarburizing action above about 1500°F. All the salts tend to adhere to the work, but this may result advantageously by protecting the surface during transfer to the quenching bath or during the cooling cycle. It may also be a disadvantage, in that some salts will not readily wash off and may, as in the case of the chlorides, cause rusting.

Active Salt Baths.—This type of salt bath is used to impregnate the surface of the charge with carbon or nitrogen in order to increase the surface hardness or to minimize scaling and decarburization.

The salts used to produce these conditions are either borax or boric oxide, B_2O_3, sodium cyanide, NaCN, or calcium cyanide, $Ca(CN)_2$, and are used in varying proportions to produce the conditions desired.

ATMOSPHERES AT HEAT-TREATING TEMPERATURES

Until recent years, comparatively little attention has been given to the enormous financial losses resulting from the loss of steel and descaling costs for the production of good, finished surfaces on such products as sheet, strip, and bar stock. To this must be added the heavy expenditures on cutting tools, etc., for the machining of badly scaled materials. Besides, losses are sustained in the engineering and heat-treating shops, where the scaled steel causes nonuniform hardness and removes metal from the surface of the heat-treated article, and this necessitates stock allowance; it frequently necessitates an extra operation, such as grinding, sand blasting, or pickling; it causes shrinkage of the steel on quenching; it clogs circulating systems of the quenching tanks; and it frequently causes burning of the steel article in grinding as a result of the loading of the grinding wheel with scale. Even persons who are unfamiliar with such working have had experience with this problem in the deterioration of oven linings, furnace grates, and electric heaters, which, in general terms, "burn out." These are, perhaps, only a few of the ways in which scale affects the over-all cost, but they are sufficient to illustrate the importance of this problem.

Within recent years certain metals and alloys have been discovered that resist heated atmospheres without scaling. What is the mechanism that enables these metals to resist certain atmospheres containing mixed gases at high temperatures? It has been determined by investigators that the resistance of metals or alloys to oxidizing environments at atmospheric or high temperatures is due to their ability to form a layer or film of protective oxide. At atmospheric temperatures this film is so thin as to be transparent to light, whereas at high temperatures it grows to a scale of visible thickness.

Oxidation Reaction.—The reaction of most interest to us in a study of this problem is

$$\text{Metal} + \text{Oxygen} \rightleftarrows \text{Metallic Oxide}$$

This is a familiar type of reaction known as reversible; *i.e.*, when it proceeds toward the right, the metal is oxidized; when it proceeds to the left, the metallic oxide is reduced. Whether this reaction proceeds in one way or the other is determined by the temperature and the pressure of the oxygen. This is merely fulfilling the workings of the law of Le Châtelier which states that "if some stress (change in temperature, pressure, or concentration) is brought to bear on a system in equilibrium, a reaction occurs, displacing the equilibrium in the direction which tends to relieve the stress." Thus, following along the same line of reasoning, there must be, for a given temperature, some pressure where the action and reaction balance. This pressure is known as the *dissociation pressure.* For further clarification of this term, let us consider the reaction

$$3Fe + 2O_2 \rightleftarrows Fe_3O_4$$

Upon heating this oxide, Fe_3O_4, *in vacuo*, *i.e.*, under conditions, if it were possible, where the oxide would no longer be in contact with the air, it would slowly start to dissociate into pure iron and oxygen. The higher the temperature of heating, the more complete would be the dissociation.

For any given temperature we find that there is a limiting amount of dissociation. This is due to the fact that as the dissociation proceeds the oxygen liberated will build up pressure inside the vessel (the O_2 pressure will now be above the dissociation pressure) and eventually equalibrium conditions will be reached and dissociation will stop. To carry this dissociation further, we have several recourses. (1) It can be done by simply increasing the temperature. This, however, will probably bring about impractically high temperatures and a second equilibrium will be set up at this high temperature. (2) The oxygen liberated can be withdrawn and the pressure maintained so low that equilibrium would never be reached, thus allowing the reaction to be completed, pure iron being left in the vessel. (3) The most practical method would be to introduce pure hydrogen into the vessel. The hydrogen unites with the oxygen to form water vapor and, if it is removed from the vessel by constant flushing with new hydrogen, the dissociation will again go to completion, provided that the temperature is held constant. If the water vapor were not removed, a new equilibrium would be established at a given temperature between O_2 and its oxide, and also between water vapor and H_2.

Equilibrium Curves.—The results of studies by physical chemists on these equilibrium conditions can be expressed by a curve, as is shown in *A* of Fig. 5-XI. In this curve the ordinate is temperature and the abscissa is the ratio of the volume percentage of water vapor present to that of hydrogen present for any given time. From the arrangement of the curve it can be seen that there are certain conditions set up that might be called ideal and others that are bad, with intermediate variations of these extremes.

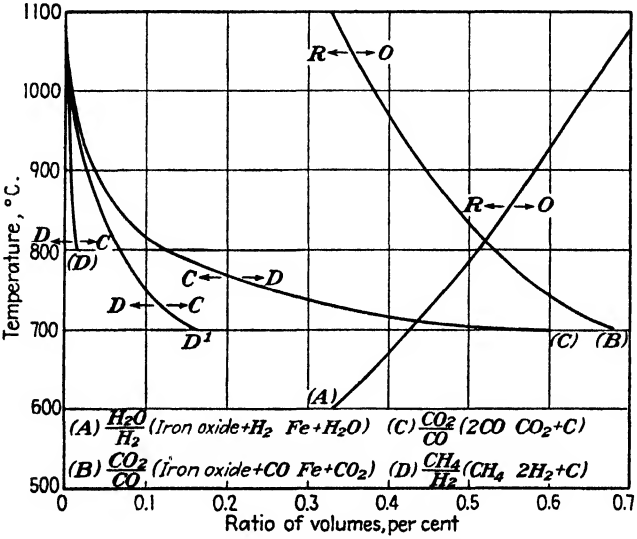

FIG. 5-XI[1].—Equilibrium curves.

If, now, a given temperature and a ratio exist such that the coordinates indicate a point to the left of the curve, the action will be reducing; *i.e.*, the reaction will proceed from left to right, the iron oxide being dissociated into Fe and H_2O in the presence

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[1] STANSEL, N. R., "Industrial Electric Heating," John Wiley & Sons, Inc., New York, 1933.

of H_2. This would, of course, postulate a previous scale formation. If the point thus formed should fall directly on the curve, the effect would be that of a neutral atmosphere and would, therefore, represent equilibrium conditions. At all points on the curve the reaction will be neither reducing nor oxidizing. These may, then, be called ideal conditions. However, if the point should fall to the right of the curve, the action would be reversed and the reaction would proceed from right to left, the water vapor combining with the Fe to form iron oxide. This would be a bad condition.

Under atmospheric conditions and within a so-called *scaling* range of temperature (750 to 2200°F.) the more common metals oxidize quite readily. From our previous discussion we can now say that, in order to protect surfaces of metals from oxidation during heat-treating cycles, the heating should be carried out in a controlled atmosphere. The most nearly ideal condition probably consists of an atmosphere neutral with respect to both oxidation and reduction, but, if there is to be any leeway, it should be toward reducing conditions. To fulfill these conditions it is necessary, then, either to have no oxygen present or to have the presence of sufficient reducing gases, such as hydrogen or carbon monoxide, to prevent oxidation.

There are some cases where the formation of scale, which we have called bad, is ideal. It has been found that the fresh-cut surface of some metals will immediately become covered by a very thin scale of extremely low permeability to oxidation which, upon forming, will cause the cessation of any further oxidation. Such films are known as *passivating* films. For example, such a film will immediately form on a freshly cut surface of aluminum at room temperature. It has been proved that oxidation has definitely ceased following the formation of the film.

Iron oxide in the presence of carbon monoxide behaves in much the same manner as it does in the presence of hydrogen, the difference between the two being the products of the reaction, which, in this case, are Fe and carbon dioxide. From curve *B*, the same conditions exist with reference to oxidation and reduction, as is shown by the direction of the arrows.

In the two previous cases only pure Fe and its oxide have been considered, but under actual conditions the situation becomes considerably more complicated, owing to the presence of carbon, which brings about several other simultaneously occurring reactions. If a piece of low carbon steel is placed in a hot furnace and is surrounded by a hydrogen atmosphere, the curve *A* would, of course, apply. At the same time, however, hydrogen will tend to react with the carbon of the steel to form methane (CH_4), until the volume ratio of methane to hydrogen and the temperature exists as a point on curve *D*, where equilibrium will be established. If this point should fall to the right of the curve, the action would be to deposit carbon, if there is methane present. This action of adding carbon to the surface of the steel is known as *carburizing*.

Carbon dioxide will react with the carbon in the steel in a manner similar to that of hydrogen. This is shown in curve *C*, where, depending upon the relation of the CO_2 to CO ratio and temperature, the reaction may either be carburizing or decarburizing, as is shown by the arrows.

With higher carbon steels, curve *D* may be shifted considerably. This is shown in curve *D′*, which represents a 0.95 per cent carbon steel.

Such equilibrium curves, showing the behavior of metals at high temperatures in the presence of gases, can be prepared for all the metals usually heat-treated and for the commonly used gases, *but the problem of practically applying this information is a difficult one. It must be remembered that a gas that would provide a protective atmosphere for one metal and a given* process would not necessarily be suitable for another metal and another process. This difficulty is due to the fact that the curves do not take into account such important factors as speed of the chemical reactions, the catalytic action of the metal whose oxide or carbide is being considered, or the other impurities occurring in the metals or on the surface of the metals.

The importance of the speed at which a reaction takes place may be illustrated by considering the following example. It is desired to anneal a low carbon strip steel at 750°C. (1382°F.) and maintain the bright finish. It is assumed that the protective gas chosen contains hydrogen and a low percentage of water vapor, so that the $H_2O:H_2$ ratio is 0.40.

From curve *A* of Fig. 5-XI it can be seen that at 750°C. (1382°F.) the conditions would be reducing; *i.e.*, if there should be any oxide present it would be reduced to water vapor and iron.

After the anneal it is necessary to cool the strip and, during the cooling, to maintain the same protective atmosphere ratio. From the curve, it can be seen that the tendency to maintain reducing conditions will become decreasingly less until a temperature of approximately 675°C. (1247°F.) is reached, at which point equilibrium conditions will exist. On further cooling, the tendency will be for the water vapor to oxidize the steel. If the rate of cooling is slow, the resulting oxidation will probably be somewhat appreciable; but, should the rate be more rapid, the amount would be slight. It must also be remembered that the type of scale formed, whether it be compact, impervious, or porous, affects the rate of penetration of oxygen and hence the speed of the reaction under given conditions.

The impurities in the steel frequently dictate the annealing procedures. Chromium steels, for example, will oxidize even at high temperatures under $H_2O:H_2$ atmosphere, the water vapor reacting with the chromium and forming a green-colored chromic oxide.

Impurities on the surface of metals are largely due to the drawing or rolling lubricants that remain on the surface after the working operations. One of the most troublesome constituents of such lubricants is the sulfur, which may react with the metal to form a sulfide and produce a troublesome sulfide discoloration.

Atmospheric Conditions.—Under atmospheric conditions there are several very important, more or less uncontrollable factors that affect the rate at which certain metals scale.

1. There is the condition where the oxide formed is so volatile at a given temperature that it evaporates from the metal surface and so permits oxidation to continue unhindered. Such metals as molybdenum and tungsten have oxides that are rather volatile at high temperatures. Take, for example, the case of an electric lamp that is broken while lighted. The oxide vapor formed by the filament is sufficiently volatile at that temperature to be swept away by the air currents, and the result is a reduced filament cross section and the retention of a bright metal surface, which may be subjected to further oxidation.

2. If the relation between the heat of oxidation of the metal and the speed of oxidation is such that the temperature of the oxidizing surfaces rises to the melting temperature of the oxide,

the oxide will run off and the reaction will continue until the metal is consumed. This happens when a thin iron sheet or an iron wire is heated to a red heat in oxygen. The metal will burn brightly because the heat of oxidation raises the temperature of the oxide to its melting point, where it drains off and affords little or no protection to the remaining metal.

3. There are the conditions involving mechanical deterioration and blistering of the scale or loosening of the scale by rapid heating or cooling.

Furnace Atmospheres.—In the previous discussion of the effects of atmospheres only the general principles have been presented. The reason for such a presentation is that the knowledge of the subject of atmospheres is at present in a stage of such rapid evolution that, besides its being beyond the scope of this text to present the details, it is also difficult to present definite conclusions without the possibility of contradiction. A few generalizations may be made, however, regarding the atmospheres in direct fuel-fired furnaces and in electric or muffle furnaces.

Fuel Furnaces.—One of the most common examples of the use of the so-called *protective* atmospheres, practiced almost universally in fuel-fired furnaces until a few years ago, was the utilization of the products of combustion,[1] the air being regulated to obtain a so-called *reducing*, *neutral*, or *oxidizing* flame. Certain heaters after long experience with a given fuel could hold a furnace near the neutral or reducing condition by watching a characteristic neutral-atmosphere haze or the smoky atmosphere of a reducing condition. Under these conditions the heater knew that the amount of scaling and the amount of decarburization (removal of carbon from the surface of the work) was materially less than when he held a sharp, live flame, which indicated more than sufficient oxygen. At their best, these methods offer only partial protection, so that if a bright surface is desired the work must be subjected to a pickling or cleaning operation after the heat-treatment.

With the recent studies on protective atmospheres, a generalization can be made regarding the effects of the individual gases

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[1] The principal constituents of combustion gases are CO_2, N_2, H_2, and water vapor, CO or O_2. The proportions of the gases present are dependent upon the degree of mixing and the relative quantities of fuel and air introduced into the burners.

as follows: Scaling or oxidizing conditions in a furnace atmosphere are produced by CO_2, O_2, SO_2, and water vapor, while the reducing conditions are produced by CO, H_2, and the unburned hydrocarbons from the fuel. The oxidizing or scaling action of the atmosphere, used at any given temperature and time of contact, is dependent upon the relative proportions of the oxidizing gases in the atmosphere. The so-called *neutral* atmosphere, obtained by burning the theoretical quantity of air and fuel, which results in the absence of O_2 and CO in the atmosphere, is also a scaling atmosphere caused by the presence of high proportions of CO_2 and water vapor. To bring about a reducing atmosphere, which materially reduces the amount of scale formed, it is necessary to burn excess fuel so that there is carried a definite amount of CO. The amount of CO necessary increases with an increase in temperature. Even with this control, however, it is practically impossible to eliminate scaling in a combustion furnace.

Cutting down the scaling by the use of reducing atmospheres, without entirely preventing it, often results in several difficulties. The first is that when thinner scales are produced they are often much tighter and more difficult to remove on pickling; and the second, that when the scale is quite thin a decarburized layer may be produced on the metal beneath it.

The soft, decarburized layer produced under certain atmosphere conditions in direct fuel-fired furnaces is found to be caused principally by moist hydrogen and water vapor. Apparently, when the scaling is heavy, the oxidation is so rapid that iron is removed from the metal as fast as the carbon; while, at a slower scaling rate, the carbon may be removed faster than the iron, and with such a speed that the carbon cannot diffuse at a rate great enough to replenish the carbon at the surface. The decarburizing effect of water vapor is due to its breakdown at the metal surface, the oxygen combining with the iron and the hydrogen with the carbon, to form CH_4. This problem may be avoided by carrying an oxidizing atmosphere with excess air; this will produce scaling but will inhibit decarburization.

The problem seems to resolve itself into an economic one in which the decision must be made whether to keep the scale at a minimum, using precaution to avoid the production of tight scale, difficult to pickle, and decarburization, and to accept the fact that pickling and grinding will be necessary; or to use definite, controlled-atmosphere methods to eliminate both scaling and decarburization.

Electric or Muffle Furnaces.—In furnaces of this type the atmosphere is normally air. Under these conditions, the furnaces are closed, to prevent drafts and restrict the amount of oxygen contacting the steel. The furnace atmospheres are not really controlled but are milder in their action with respect to the amount of scale formed. A heavy, free scale will be formed at temperatures above approximately 1200°F., with the amount increasing greatly with an increase in temperature.

The developments in maintaining bright heating have been confined principally to this type, owing to the fact that it has been found impractical to operate direct fuel-fired furnaces with atmospheres containing a sufficient amount of CO and H_2 to counterbalance the scaling action of CO_2 and water vapor. In order to accomplish absolute bright heating, it is necessary to use a gas that will provide a neutral atmosphere, so that there will be no scaling or decarburizing at the temperature and time required for the heating treatment. It must be noted that the only truly neutral atmospheres are those using the rare gases, but these are applied only in special cases because of their cost. All other furnace atmospheres must be regarded as chemically active, the action being dependent upon the composition of the metal, the composition of the atmosphere, and the heating temperature. The most common atmospheres employed include those resulting from the partial or complete combustion of coal, gas, or oil; hydrogen from electrolytic dissociation of water; and dissociated ammonia.

Suggested Questions for Study and Class Discussion

1. What would be considered the entire cycle of a heat-treating cycle?
2. Compare the principles involved in heating and those involved in cooling.
3. Describe the effects of heat-treating varied shapes and sizes of objects.
4. Why should good pyrometric control be maintained on the entire heat treating cycle?
5. What are the three ways for arranging the heating chamber in hearth furnaces?
6. What are the advantages of the batch- and continuous-type furnaces in the heat-treatment of steel?

7. What testing procedure would you follow in determining whether any given material being heated with a furnace is being heated to the desired temperature?

8. What effects would be produced in steel by water quenching at various temperatures? Discuss this with reference to the typical cooling curve.

9. What are the benefits of oil over water in quenching? What other method do we have for producing the same effect? Explain.

10. Discuss the precautions that are necessary in obtaining full benefit of the quench.

11. What type of furnaces would be most satisfactory for maintaining controlled atmospheres? Why?

12. What atmosphere would you suggest for bright-annealing strip steel or wire?

13. From Fig. 5-XI consider only curves *B* and *C*. We are annealing 1 in. round bars ground free of scale and decarburization prior to annealing. The annealing temperature is 900°C.

a. Would the steel scale while at heat in the furnace if we maintained a CO_2:CO ratio of 0.5?

b. Under the same conditions with a ratio of 0.4?

c. Would the steel decarburize under condition (*a*)?

d. Would the steel decarburize under condition (*b*)?

e. In annealing bright steel free of scale or decarburization prior to annealing, approximately what ratio of CO_2 would be necessary to prevent scaling, carburization, or decarburization while at heat, when the annealing temperature is (1) 900°C., (2) 800°C., and (3) 700°C.?

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CHAPTER XII

PRECIPITATION HARDENING

In the previous discussion of heat-treatment it was considered that the low carbon steels were not materially affected by quenching and tempering as therein described. If, however, a very low carbon steel of, say, 0.05 per cent C. is quenched not from above the upper critical, as is the usual practice, but from below it, about 1300°F., and then tempered at 110°F. for approximately 10 hr., an interesting result will be obtained. In the as-quenched condition, the Brinell hardness will perhaps be 120, while in the tempered condition it will be approximately 140. If this same steel is allowed to stand for a month or more at room temperature, its Brinell hardness will increase to as much as 170. This change may also be observed in the higher carbon steels, but as the carbon is increased the phenomenon is less readily observed, because of the increasingly higher hardness values.

MECHANISM OF PRECIPITATION HARDENING OR AGING

The results of many observations, such as the above, on both ferrous and nonferrous materials that show this phenomenon, have brought about the deduction that the behavior of the steel is like that of the precipitation-hardening alloys in general. The hardening effect under the conditions cited takes place so slowly that the process is usually termed *aging* or *age hardening* rather than tempering, since tempering usually denotes a degree of softening. The general explanation for these phenomena may be explained by use of the typical constitutional diagram, Fig. 10-V, which illustrates the solid-solubility characteristics necessary for age hardening. This diagram may be recognized from previous studies to be of the partial solubility in the solid-state type. In the alloys represented by this binary diagram, a solute *B*, generally a compound, has a considerably higher solubility in the matrix *A* at higher than at lower temperatures. If we were to quench from the higher temperature, the result would be a supersaturated solid solution of B in A. If this solid solution is allowed to stand for a long period of time or if, as is more commonly practiced, it is heated, the hardness will increase; while on more prolonged or higher heating the hardness will tend to decrease. This phenomenon is explained as resulting from the precipitation of the finely dispersed particles from the supersaturated solid solution. On further heating, the particles tend to grow large and, as a result, have a less drastic hardening effect. With the data available at the present time, this explanation is substantially correct. There are, however, certain complexities inherent in precipitation in the solid state, but these will not be discussed.

In the case of the iron-carbon system, the constituent B is iron carbide, which, on standing at room temperature, may be precipitated in particles too small to be observed with the most powerful microscope, but which greatly changes the mechanical properties, such as strength, hardness, etc.

According to Davenport and Bain,[1] aging is a process in which the trend is toward a restoration of real equilibrium, constitutional or mechanical, and away from an unstable condition set up by a previous final operation—*e.g.*, cooling abruptly from other temperature ranges—or a mechanical treatment introducing stresses. The rate of approach to equilibrium of a system forced into an unstable state is always time consuming, the actual time varying from a fraction of a second to years, depending upon the conditions.

AGING IN STEEL

Causes.—Prior to the discussion on the aging characteristics of particular steels, it should be stated again that aging in steel is primarily a precipitation phenomenon and may result from any of the following causes:[2]

Allotropism.—The transformation of one lattice to another usually occurs with a change in solid solubility. This is particularly true in alloys that transform from body-centered to face-centered cubic lattices, since the face-centered structure possesses

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[1] Davenport and Bain, The Aging of Steel, *Trans. A.S.M.*, **23**, 1047–1106 (1935).

[2] Shapiro, Carl L., Color Carbon and Aging, *Trans. A.S.M.*, **27**, 666–696 (1939).

a greater dissolving power than that of any other polymorphic form of the material. This especially pertains to steel, since austenite (face-centered cubic) can retain approximately 1.7 per cent carbon in solution, while ferrite (body-centered cubic) can dissolve only a maximum of approximately 0.035 per cent at the A_1 point or about 0.006 per cent at room temperature.

Limited Solubility.—The solid solubility of carbon in alpha iron decreases progressively from approximately 0.035 per cent at the lower limit of the thermal critical range A_1 to some 0.006 per cent at room temperature.

Mechanical Deformation.—Cold work will usually cause aging by shifting the solid solubility slightly to the left, thereby causing precipitation and subsequent aging.

Diffusion.—Precipitation aging may occur by the heating of an alloy in contact with a substance—a gas, metal, compound or solid solution—that diffuses into the alloy and affects the solid solubility. Consequently, when the metal is cooled to room temperature, supersaturation and precipitation result in aging. This occurs primarily in carburizing, nitriding, or any process of impregnation where the solubility at elevated temperature is greater than at room temperature.

Controlling Factors.—The above-mentioned factors control the affinity of the steel, heat-treated in any condition, to age. Therefore, aging in steel is controlled by (1) the amount of retained austenite, (2) the amount of carbon dissolved in alpha iron, (3) the size of the iron-carbide particles, (4) the time and temperature of aging or tempering, and (5) the degree of mechanical deformation remaining within or conferred upon the steel after the final treatment.

HARDNESS OF DECOMPOSITION PRODUCTS

The results and the theory regarding the quenching and tempering of carbon steels have been discussed in the previous section. The results, briefly summarized, are as follows: The initial step in the austenitic decomposition on full quenching is the lattice change from face-centered cubic to body-centered cubic. This transformation is accompanied by supersaturation in which the carbon is retained briefly in solution in the ferrite before precipitation occurs. At the first instant of precipitation the carbide particles are precipitated in a submicroscopic state of division and further agglomeration of the particles will occur as rapidly as the conditions of time and temperature will permit.

The tempering of hardened steels is considered to take place in three steps.[1] (1) At a temperature of approximately 210°F. tetragonal martensite changes to cubic and this change is accompanied by the precipitation of carbides out of the solid solution. (2) At approximately 450°F. the remaining carbide is precipitated from the martensite and any austenite that might be retained breaks down to martensite, which in turn is tempered to the same stage as the martensite formed on quenching. (3) At about 570°F. the reaction increases in velocity so that discrete carbide particles will be formed, which begin to coalesce.

It has previously been stated that martensite is the hardest of the decomposition products formed on quenching. The drastic change in hardness produced on quenching has made this phenomenon the point of considerable speculation and research, in an effort to find a cause. The result of these studies, even though there are certain complexities that tend to make definite conclusions regarding the hardness very difficult, is to explain the martensitic hardness as due to the precipitation of carbon or carbide from the supersaturated solid solution. This explanation has been arrived at by analogy with the hardening effects of the many other age-hardening alloys. As the possible cause for this hardness, many other factors have been suggested, *e.g.*, the allotropic change from gamma to alpha, the tetragonal structure, the acicular structure, the fine grain size, the distortion of the lattice, and the presence of internal strains; but these are considered subordinate factors.

AGING OF LOW CARBON STEELS

The term "aging" as applied to soft, or low carbon, steels, relates to a wide variety of commercially important, slow, gradual changes that take place in properties of steels after the final treatment. These changes, which bring about a condition of increased hardness, elastic limit, and tensile strength, with a consequent loss in ductility, occur during the period in which the steel is at normal temperatures. This phenomenon has been

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[1] EPSTEIN, SAMUEL, "The Alloys of Iron and Carbon," Vol. I, Constitution, pp. 233–234, McGraw-Hill Book Company, Inc., New York, 1936.

observed in metals and alloys both after quenching and after cold deformation.

The aging effect in low carbon steels is the result of two phenomena: (1) quench aging, and (2) strain aging. Quench aging may be defined as the term applied to those changes that take place in steel following a final operation consisting of fairly rapid cooling from an elevated temperature. *Strain aging* may be defined as the term applied to those changes that take place in steel when the final operation consists of cold working.

Quench Aging.—This type of aging is most frequently observed in annealed wire or sheet of low carbon content of about 0.04 to 0.12 per cent, which, immediately after the treatment, will possess low hardness and high ductility; but which, after a few weeks storage, is found to have become harder and to have lost some of its ductility. It must be stressed that this type of aging is restricted to annealed metal, free from cold work, which has been cooled rather rapidly in the range below 1100°F. (600°C.).

This type of aging is believed to be an example of the precipitation-hardening theory previously discussed. According to the principle as therein described, it was found that a solid solution supersaturated by rapid cooling from some high temperature subsequently develops, at some lower temperature, a shower of particles of the solute metal or compounds, in release of this supersaturation. This precipitation may occur at temperatures as high as 1100°F. (600°C.) or below room temperature, depending upon the alloy.

In the previous discussion of heat-treating it was convenient to neglect the slight solubility of carbon in iron at low temperatures but now, by comparing the enlarged low carbon portion of the iron-carbon diagram, Fig. 1-XII, with the typical constitutional diagram illustrating the solid-solubility characteristics necessary for age hardening, we see that the iron-carbon alloys, with only a few thousandths per cent of carbon when heated not above 1300°F. (700°C.), have the proper requisites for age hardening. Hence, this slight solubility becomes extremely important in the consideration of age hardening.

Let us consider, for example, a steel of 0.06 per cent carbon slowly cooled to a temperature of about 700°C. (1290°F.), where it consists of an aggregate of ferrite and pearlite. The ferrite, according to the iron-carbon diagram, may retain some carbon in solid solution and, if a study were made of the Fe-O and the Fe-N systems, it would be found that ferrite may also retain some oxygen and some nitrogen along with the carbon in solid solution, since these diagrams also denote solid solubility in the solid state. In all cases the solid solubility decreases with falling temperature. Therefore, it should be possible to retain ferrite in the supersaturated condition by quenching, provided that it contains more of these elements, or at least more of one of them, than is normally soluble at room temperature.

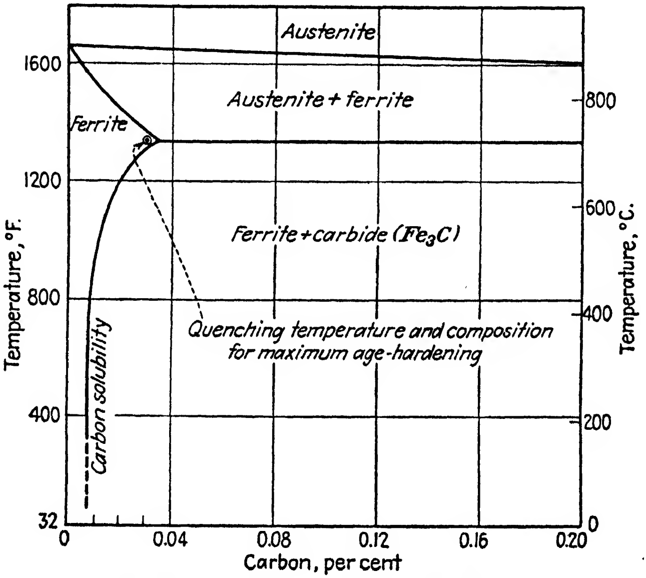

FIG. 1-XII.—Enlarged low carbon end of the iron-carbon equilibrium diagram. This serves to emphasize the solubility of carbon in alpha iron and the age-hardening potentialities of the system.

These considerations make it evident that ferrite is the aging constituent and that should iron-carbon alloys be quenched from around 700°C. (1290°F.), where ferrite has the maximum solubility for carbon, oxygen, and nitrogen, a supersaturated solution should be obtained, which should age. The more ferrite present, then—*i.e.*, the less carbon in the alloy—the more it should age after quenching, provided that the above conditions are evident. Aging by quenching below the critical differs in this respect from aging after quenching from above the range, in which case the more carbon there is present the more intense is the effect.

From Fig. 2-XII,[1] we note the effect of the rate of aging of a 0.06 per cent carbon steel after quenching from 720°C. (1325°F.), with respect to increased reheating temperatures or aging temperatures. The data have been plotted on semilogarithmic paper with Rockwell B hardness as the ordinate and time on the logarithmic scale as the abscissa. In general, it can be seen that at low temperatures the hardening effect is somewhat slow, but that it becomes rather rapid, although less intense, at higher temperatures. Further, it can be seen that reheating in some definite temperature range causes first a hardening and then a softening of the metal. For example, at 105°F. (40°C.), a maximum hardness is reached at 200 hr. of aging, which on continued heating at this temperature starts gradually to decrease. Similarly, upon reheating at 175°F. (80°C.), a hardness maximum is reached at 2 hr., which within 200 hr. has decreased to essentially the original hardness. With higher temperature, the steel passes through the hardening cycle and regains its softness with great rapidity. This circumstance of softening after some precipitation hardening is due to the coalescence of the carbide particles to form particles too large to act as effective hardening agents. Apparently, all the carbon in excess of the solubility limit at that temperature is rejected,

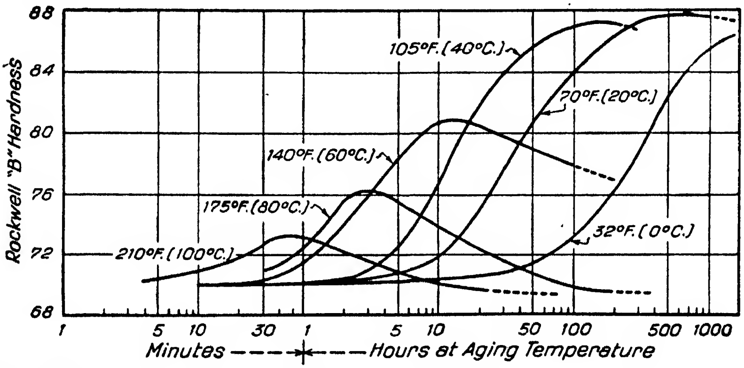

FIG. 2-XII.—Quench aging of 0.06 per cent carbon steel. Plotted on logarithmic time scale. (*Courtesy of American Society for Metals.*)

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[1] DAVENPORT and BAIN, The Aging of Steel, *Trans. A.S.M.*, **23** (No. 4) (1935).

the resulting structure possibly being softer than the as-quenched structure.

If the treatment thus described has resulted in the rejection of essentially all the dissolved carbon and has been sufficient to produce a dispersion of no marked hardening power, the metal will then be in a state that will not change substantially thereafter. The metal in this state is spoken of as *preaged* or *stabilized.*

The quench-aging effect on mild steels with respect to the method of deoxidation is summarized by Daniloff, Mehl and Herty[1] as follows:

> 1. The susceptibility of normalized rimmed steels to quench aging is very high, the maximum being reached in steels containing 0.04 to 0.05 per cent carbon. In the semi-killed steels this aging is considerably less than in rimmed steels, and the susceptibility to this aging shows the same general tendency to diminish with increasing carbon content. The silicon-killed steels quench-age considerably but only about one-half as much as the rimmed steels.
>
> 2. Steels killed with aluminum quench-age very little, the degree decreasing with increasing carbon content. In the carbon range of from 0.12 to 0.25 per cent they become practically non-quench-aging.

Deoxidation, it is stated, decreases the tendency of mild steel to age because (1) it lowers the dissolved oxygen content of the steel, (2) through the formation of finely divided particles of deoxidation product it refines the grain size (aluminum-killed), and (3) it, possibly, reduces the rate of carbon diffusion in ferrite.

Strain Aging.—This type of aging occurs in steels that have been subjected to cold-work deformation, *e.g.*, cold rolling or wire drawing. Cold work will, of course, cause the usual increase in hardness, but it is observed that most low carbon steels will continue to lose some of their apparent ductility for some time following the cold-working operation. This action proceeds quite slowly at room temperature, but quite rapidly at higher temperatures or just above room temperature. To induce this type of aging, it is not necessary that solid solutions be obtained in a supersaturated condition through quenching. The cold working of steels in the annealed-condition and, hence, not in a supersaturated solution may be followed by aging.

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[1] DANILOFF, MEHL, and HERTY, The Influence of Deoxidation on the Aging of Mild Steels, *Trans. A.S.M.*, **24**, 595–639 (1936).

The Nature of Strain Aging.—The characteristics observed in this type of aging suggest that this phenomenon is a type of precipitation reaction. Sauveur[1] and others say that cold-work deformation causes a distortion of the space lattice of the iron, in which condition it is unable to retain much of the solute in solution, the distorted lattice acting in this respect like a supersaturated solution. The term "solute" has been used, owing to the fact that there is still insufficient evidence on the exact nature of the precipitating compound. Present evidence[2] seems to indicate that the oxygen in the metal is the element that combines with the iron, forming an iron-oxygen compound that is rejected in the slip bands of the cold-worked grains. This type of aging is, therefore, not confined to solid solutions capable of being obtained in the supersaturated condition by quenching. Some solid solutions should be subject to age hardening even though there is no decreasing solubility with fall of temperature. This, therefore, indicates the difference between aging after quenching and aging after cold-work deformation.

The "urge to age" resulting from the cold-work distortion may be so great that the rejection of the compound may take place during working operation. This assumption affords an explanation of the "blue-brittleness" phenomenon, which is the peculiar hardening that takes place in rolled or drawn soft steel when the metal in process reaches a temperature of around 400 to 600°F. It has been called *blue brittleness* because of the surface oxidation that takes place at that temperature, producing a characteristic blue color (temper color). From this, then (should there be noticed little or no aging following cold work) it can be deduced that there has been considerable precipitation during the operation, leaving little of the compound for subsequent precipitation. It has been noted also that a slight amount of cold-work deformation is generally followed by more aging than in the case of metals more severely worked.

Characteristics.—Some of the outstanding characteristics of strain aging as summarized by Davenport and Bain[3] are (1) that

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[1] Sauveur, Albert, Notes on the Aging of Metals and Alloys, *Trans. A.S.S.T.*, **22**, 97–119 (1934).

[2] Davenport and Bain, The Aging of Steel, *Trans. A.S.M.*, **23**, 1047–1106 (1935).

[3] Davenport and Bain, The Aging of Steel, *Trans. A.S.M.*, **23**, 1070 (1935).

the maximum hardness is not so definitely the result of low temperature and long time as in quench aging, but that on the contrary, within wide limits, the higher the aging temperature, the greater is the hardening; (2) that softening by "preaging" does not occur the same as in quench aging, but that, instead, a wide range of temperatures will cause practically maximum hardening without subsequent softening even after long time intervals; and (3) that hardening is quite rapid even at very low temperatures.

Other Evidences of Strain Aging—Intergranular Corrosion.—It has been stated previously that the slip planes in the grains constitute points for precipitation and, by analogy, it should seem very possible that in annealed or slightly cold-worked metal, the grain boundaries should likewise constitute suitable points for the compound precipitation. We should, therefore, expect that such a situation would be evidenced by the loss of intergranular ductility or by some form of intergranular corrosive attack. Both of these effects are observed in metals that are susceptible to strain aging. Without a detailed explanation, let us take the one example of the so-called *caustic*, or *alkali*, *embrittlement* that occurs in boilers. It seems that in certain positions in the boiler, around joints or punched holes where there has been a little cold work, the modern operating temperatures are sufficient to bring about rapid aging. The alkali salts in the boiler feed water, which may easily concentrate in these isolated positions, may in time produce an intergranular corrosive attack by reaction of the salts with the oxide particles. The removal of this dispersion of oxide will cause a rupture of the metal along the grain boundaries; the stress causing rupture probably is effected by the steam pressure. This effect is shown in Fig. 3-XII (15).

Most low carbon steels are somewhat subject to caustic embrittlement, although in boiler construction the severity of the attack is minimized by removing the moderately cold-worked metal from the vicinity of the rivet holes, etc., or by employing a steel that is immune to such attack; *e.g.*, a 12 per cent chromium steel.

Stretcher Strains.—Another very interesting change in plastic properties is evidenced when annealed low carbon steel is undergoing deformation. If a strip of such steel is placed in a tensile

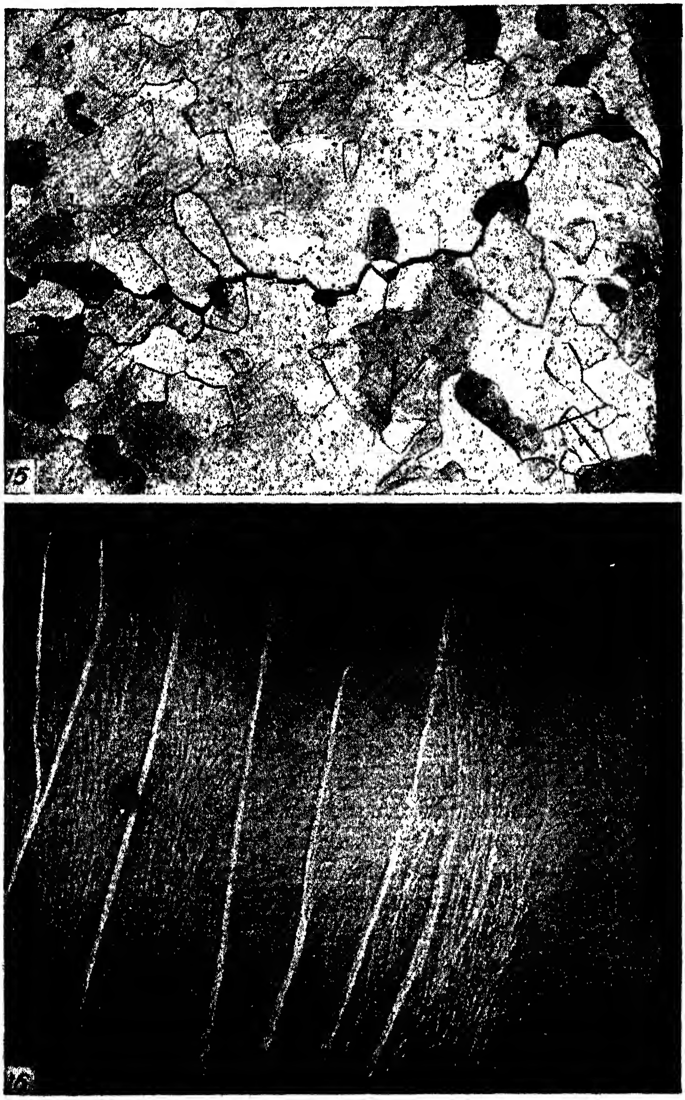

FIG. 3-XII.—15, intergranular corrosion crack in specimen maintained under elastic stress in strong alkali salt solution at about 210°F. (100°C.). × 125. 16, typical stretcher strains in an automobile stamping. (*Courtesy of American Society for Metals.*)

machine and stressed in tension up to the limit of elasticity, it is found to yield and elongate in one or two sections as much as 10 per cent. The load that the specimen was able to carry before this sudden elongation or yield may also be lowered as much as 15 per cent. This process of plastic deformation is best postulated by considering that it takes place in steps. For example, assume that there are three grains in such an arrangement in a sheet that one grain will yield at a stress of 25,000 p.s.i., another at 26,000, and the last at 27,000. Upon application of the stress below 25,000, the deformation will be purely elastic but, after reaching that figure, the resistance to slip will be overcome and there will be a yielding of a grain with the formation of a so-called *slip band.* This momentary yielding reduces the load and the other two grains will return to their original shape. Upon further stress application, the intensity of this stress on the yielded grain is somewhat greater because of its reduced cross section; but as a result of work hardening, the resistance to slip is also increased. On reaching 26,000 p.s.i., the second grain will yield, form a slip band, and work harden. The third grain will follow the same procedure of slipping and hardening. If these alternate slippings and work hardenings were recorded on a stress-strain curve, they would be evidenced by a series of fluctuations, after which the curve would rise again as a relatively smooth curve.

This phenomenon of sudden local yielding may be noted in bending an annealed soft-iron wire, but it is best portrayed in deep-drawing or stamping operations. These zones of sudden elongation are named after their discoverer, Luder's lines. In deep-drawing and stamping operations, which seem particularly suited to the development of these bands in sheet or strip metal, the names "stretcher strains" and "worms" have been applied. A typical example of stretcher strains is shown in the automobile body section, Fig. 3-XII (16).

In many applications this surface condition is serious, but fortunately the remedy is simple. It is accomplished by the application of cold work, even in a small amount, which causes a small but definite slip in each grain, places the metal in the second hardening condition, and causes the disappearance of the sudden local yielding and the stretcher strains. The effect of the application of increasing amounts of cold work on the yield point is shown in the series of stress-strain diagrams of Fig. 4-XII. This work is effected commercially by stretcher leveling, a pinch pass, or roller leveling. There is no appreciable hardening from such work, nor is there any appreciable change in the gauge of the plate or strip. The break at the yield point in the stress-strain curve with 0 per cent cold reduction may be made to disappear not only by cold work but also by quenching from about 1290°F.

The effect of these light treatments is not permanent and will, in the case of ordinary low carbon steel, revert to the stretcher straining or "annealed" condition, except when the amount of cold work applied to prevent the condition is so severe that undesirable hardening is produced. The rate at which this condition is regained is apparently inversely proportional to the amount of cold-work deformation and to the temperature of the metal during the interval following the light cold working. The increased susceptibility to stretcher straining after light cold working is illustrated in the upper half of Fig. 5-XII. This figure shows the effect of aging on ordinary mild steel, deep-drawing sheet at room temperature, after cold rolling 1 per cent. With the same percentage of cold work and an aging temperature of 400°F., the time interval to produce a degree of susceptibility to stretcher straining approximately equivalent to that shown for 1 year in Fig. 5-XII would be 1 min.

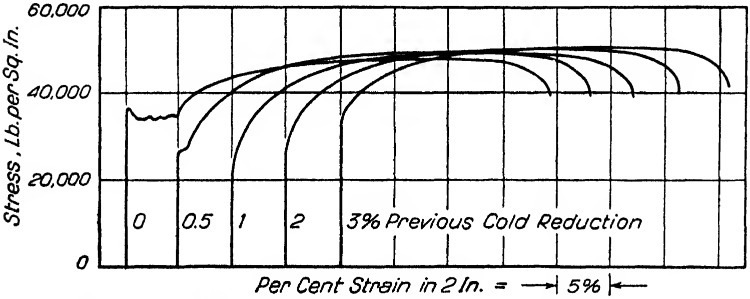

FIG. 4-XII.—Stress-strain curves of mild steel sheets subjected to the indicated amounts of cold work prior to tensile testing. Note the absence of marked yield point in materials subjected to about 1 per cent or more cold reduction. (*Courtesy of American Society for Metals.*)

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[1] HAYES and GRIFFIS, Deep Drawing, *Metals and Alloys*, **5**, 110–112 (1934).

The change in shape of the stress-strain curves and the return of the sharp yield point with elapsed time is evident. The return of the yield point is accompanied by stretcher straining of the material.

The stress-strain curves in the lower half of Fig. 5-XII are of a series of samples of so-called stabilized material after cold rolling 1 per cent and aging for various lengths of time at room temperature. In this material, it is evident that the aging has not affected the shape of the stress-strain curve. There is no return to the sharp yield point and the material will remain free from stretcher strains.

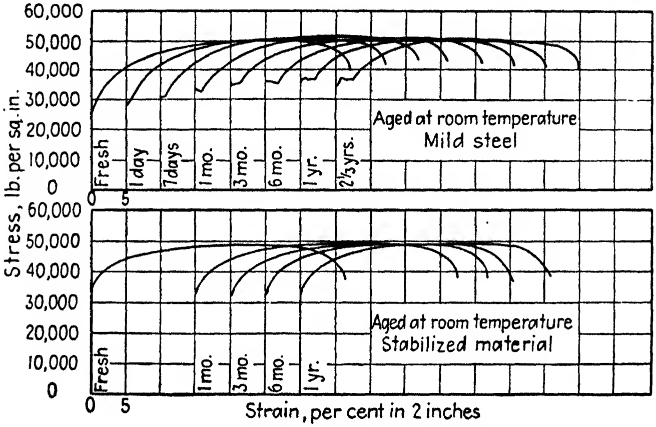

FIG. 5-XII.—(Upper) Stress-strain curves of ordinary mild steel, cold rolled 1 per cent, and aged at room temperature. (Lower) Same for stabilized material. Typical composition of stabilized or nonaging material strongly deoxidized with aluminum and titanium.

C %	Mn %	P %	Si %	Ti %	Al %
0.05	0.03	0.005	.....	0.067	0.042
0.042	0.47	0.011	0.073	0.058	0.037

Production of the so-called stabilized material involves the attainment of the desired analysis of steel, using a strong deoxidizer, such as titanium, aluminum, etc., followed by a stabilizing heat-treatment. For the analyses shown under Fig. 5-XII a suggested treatment[1] for deep-drawing stock is to normalize and then reheat to 1180°F. for approximately 3 hr., followed by cooling at a rate of 10 to 15°F. per hr.

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[1] DAVENPORT and BAIN, The Aging of Steel, *Trans. A.S.M.*, **23**, 1047–1106 (1935).

The rate of recovery of the marked yield point in the mild steel following cold work is quite similar to the strain-aging phenomenon of the same steel, but as yet this relation is rather problematical. It is significant, however, that those steels that have been deoxidized in such a manner as to eliminate substantially all the age hardening after cold work are also practically free from the return of stretcher-strain behavior. This, by analogy, would lead one to believe that there is some possible connection between the return of stretcher straining and the hardening by strain aging.

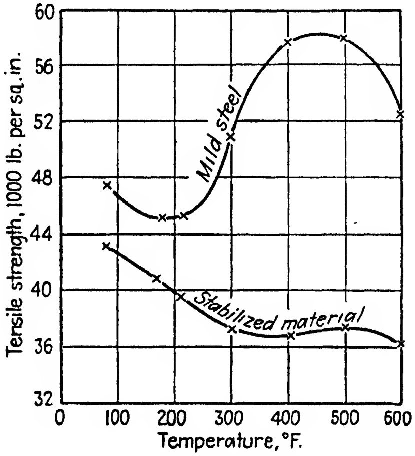

Fig. 6-XII.—Blue brittle characteristics of mild steel and stabilized material as shown by the effect of temperature on tensile strength.

Blue Brittleness.—The material that is stable against changes due to aging is also free from the blue-brittleness phenomenon previously mentioned. This is indicated by the curves in Fig. 6-XII,[1] which shows the effects of testing temperature on the tensile strength of mild-steel and stabilized-steel sheets. As is evident, the tensile strength for the mild steel decreases to about 212°F. and then increases, reaching its maximum in the neighborhood of 500°F. (blue-heat range); while, for the stabilized material the strength decreases continuously with rising temperature. Kenyon states that the tensile test in the blue-heat range is one

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[1] Kenyon and Burns, Testing Sheet for Blue Brittleness and Stability against Changes Due to Aging, *Proc. A.S.T.M.*, **34** (Part 2), 48–58 (1934).

of the simplest ways to distinguish between aging and nonaging material.

Suggested Questions for Study and Class Discussion

1. Why is steel hard?

2. What connection does the theory that you have advanced have with the quench- and strain-aging phenomena? How does it differ?

3. What effect does steel practice have on quench aging?

4. What is the effect of carbon in bringing about hardness increases in quenching from above the critical range and in quenching below the critical range? Explain.

5. What practical significance might be developed from the aging phenomena in rolled shapes or cast-steel products of thin and intricate shapes?

6. Explain the phenomenon of stretcher straining. In what way is this connected with aging?

7. A certain automobile manufacturer placed an order with a steel company for cold-drawn seamless tubing with the following specifications:

Analysis	C, 12–15 per cent; Mn, 30–60 per cent; P, 0.30 per cent; S, 0.40 per cent
Type of steel	Killed
Finish size	1$\frac{5}{8}$ × .065 in. Wall
Hardness	90 Rockwell B, minimum.

It was felt by the steel company that the hardness and surface finish could be obtained by quenching the hot-rolled pipe and finishing with a couple of cold passes on a draw bench, thus eliminating the necessity of intermediate anneals during cold drawing. When the pipe was tested a short time after cold drawing, it was found that the desired minimum in hardness could not be obtained. After standing for several days, however, the pipe was found to be in the specification range. The following hardness figures will illustrate.

Process	Rockwell B
Hot-rolled quench	75B
Cold-draw pass	87B
After standing	95B

a. Explain the reason for the observed variation in hardness figures.

b. Assuming the same above specifications with the exception that we are now interested in surface appearance with maximum softness and ductility, what processing would you suggest?

c. What would be the final microstructure that would give maximum softness and ductility? Let us assume a lower critical temperature of 1360°F.

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CHAPTER XIII

GRAIN SIZE IN STEEL

In all our previous studies we have observed that there is a definite relationship between the properties of the piece of metal that is being tested and the size of the crystals that compose the piece. Since we have also observed that the size of these grains may be changed by heat-treatment, it becomes important for us to inquire into the laws governing the change of grain size. It must be stressed at this point that by grain size in this discussion, we do not mean ferrite grain size, as in a low carbon steel, or the size of the roughly triangular patches of pearlite that are found in a medium carbon steel. In eutectoid pearlite it is difficult to determine just what might constitute a grain. In this discussion we will consider the austenitic grain size which is established above the critical transformation range.

In the low carbon steels we should have an understanding of the manner in which the ferrite grain size changes when the metal is cold-worked and then heated. In particular, we have found that with a certain moderate amount of cold work, followed by heating to a certain temperature range, abnormally large grains were formed and that, with different amounts of cold work and different heating temperatures, a whole range of grain sizes could be produced.

It has also been stressed that when steels are heated above their transformation range the grains of austenite will grow and that, the higher the heating temperature, the larger the grains will become. It is well known now that certain steels will coarsen quite readily with overheating, while others will show definite resistance toward coarsening. Steels that exhibit a marked increase in grain size on heating within the temperature limits for commercial heat-treating operations have been called *coarse-grained;* while steels characterized by but little grain growth at these temperatures are termed *fine-grained.* There may occur also a mixed size (referred to as *duplexed*), in which there is a combination of fine and coarse grains. This condition merely indicates the natural transition stage from fine to coarse grains.

METHODS OF ESTIMATING GRAIN SIZE

The most commonly accepted method of grading heats of steels is by chemical analysis but, since it has been found that all heats of similar analysis do not act alike, several other methods of grading have been developed.

Microscopic Examination. *McQuaid-Ehn Test.*—The McQuaid-Ehn test, which was developed in 1922, was the original means for determination of the austenitic grain size and, as originally developed, was intended for use on plain carbon, case-hardening steels. This test consists in the measurement of the size of the austenite grains that constitute the structure after a sample has been heated for 8 hr. at a temperature of 1700°F. At this temperature many difficulties are encountered in attempting a direct measurement, so it was necessary to devise some means of tracing, in the steel at ordinary temperatures, the austenitic grain size that has existed at the end of the heating period. The solution was found by heating the sample in an active carburizing medium for 8 hr. at 1700°F. and cooling it very slowly. The absorption of carbon by this treatment is sufficient to increase the carbon content of the outer edge or case and thereby produces a hypereutectoid area.

Austenitic Grain Size.—The steel at the carburizing temperature (1700°F.) is, as we know, austenitic and the grains are, therefore, known as *austenite grains.* During carburizing, the austenitic grains at the surface of the sample will become saturated with carbon and, on slow cooling, will no longer be able to retain all the dissolved carbon in solution so that the excess is precipitated as cementite, which gathers at the boundaries of the austenite grains to form a cementite network around them. The final microstructure in the hypereutectoid area, then, is a pearlitic ground mass with a network of proeutectoid cementite (Fig. 1-XIII). The McQuaid-Ehn grain size is based on a study of the size and number of the austenite grains as shown by the cementite markings in the hypereutectoid area.

The application of this test as a universal method has been shown to fall short for several reasons.[1]

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[1] Vilella and Bain, Revealing the Austenitic Grain Size of Steel, *Metal Progress,* **30** (No. 3), 39 (1936).

1. It fails to show the really significant grain size, *i.e.*, that established at the heating temperature employed in practice (and responsible for the steel's properties), which is not necessarily that produced at the otherwise irrelevant temperature of 1700°F.
2. By introducing carbon and oxygen it can disclose only the grain size of the steel as thus possibly modified, and this size is frequently not the same as that of the unaltered metal.
3. The entire test is too time consuming.

Normality.—The McQuaid-Ehn grain-size test, besides showing the austenitic grain size, may be used to indicate what is termed *normality*. The terms "normal" and "abnormal" are used to differentiate between two extremes in structure. A "normal" structure is one in which the grains are outlined by a continuous cementite network of fairly regular width, with the internal structure showing lamellar pearlite (Figs. 1-XIII and 2-XIII). An abnormal structure is one in which the pearlite, if any is formed, is coarse and irregularly lamellar and is completely broken down at the grain boundaries to form massive cementite and free ferrite. This breakdown may appear within the grains, also. The thickness of the network is usually variable and sometimes discontinuous (Figs. 3-XIII and 4-XIII).

These terms, as originally applied to steels, are somewhat misleading and have caused considerable confusion. Since it has been found that normality and abnormality can exist in both fine- and coarse-grained steels and that the use of these names does not imply that a steel is of inferior or questionable quality, since each type is applicable in certain fields, the terms properly pertain to a structure and not to a steel.

Other Methods.—Grain size in given specimens of steel may be estimated microscopically in other ways, as well. These methods are all dependent upon the method of cooling the steel from the austenitic condition. The procedure to be used is found to vary with the type of steel and the carbon content.[1]

In most cases of carbon steels, where the carbon content ranges between 0.25 to 0.60 per cent, it is found sufficient to heat a specimen about ½ in. in diameter at the desired temperature for sufficient time, after which it is removed and cooled in still air (normalized). Specimens thus treated will show the austenite

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[1] Bain and Vilella, Austenitic Grain Size in Steel, A.S.M. "Metals Handbook," pp. 754–767, Cleveland, Ohio, 1939.

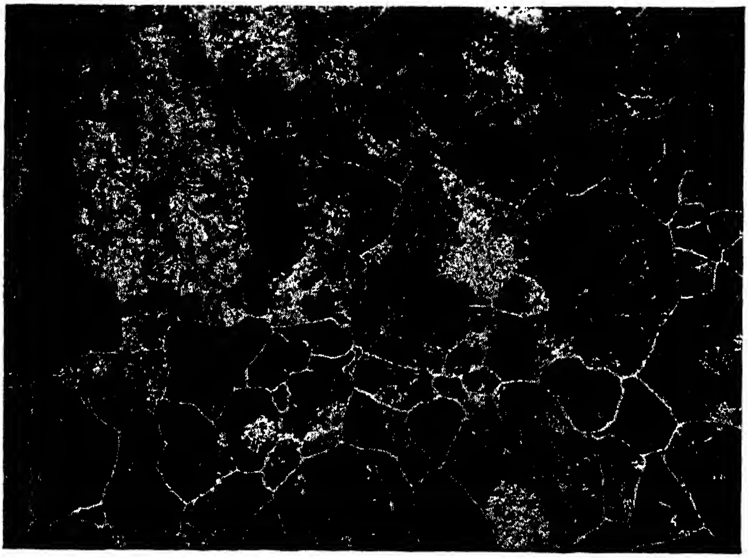

Fig. 1-XIII.—McQuaid-Ehn test. Coarse-grained normal steel. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

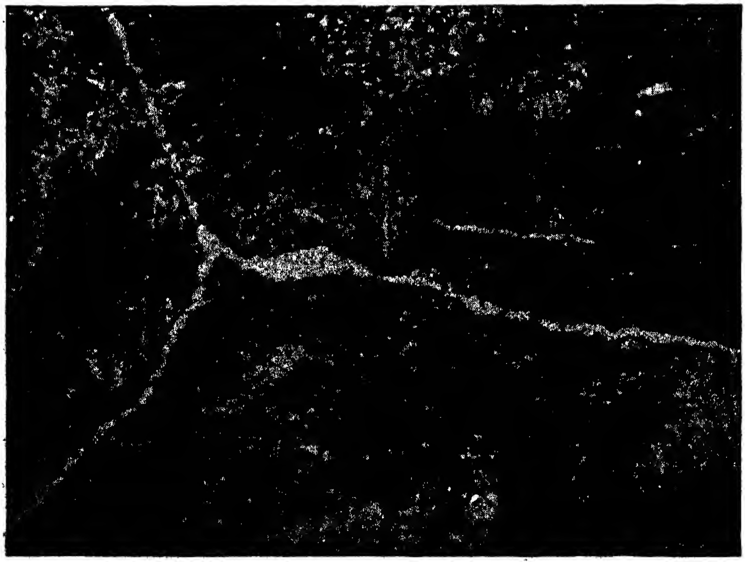

Fig. 2-XIII.—McQuaid-Ehn test. Coarse-grained normal steel. × 1,000. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

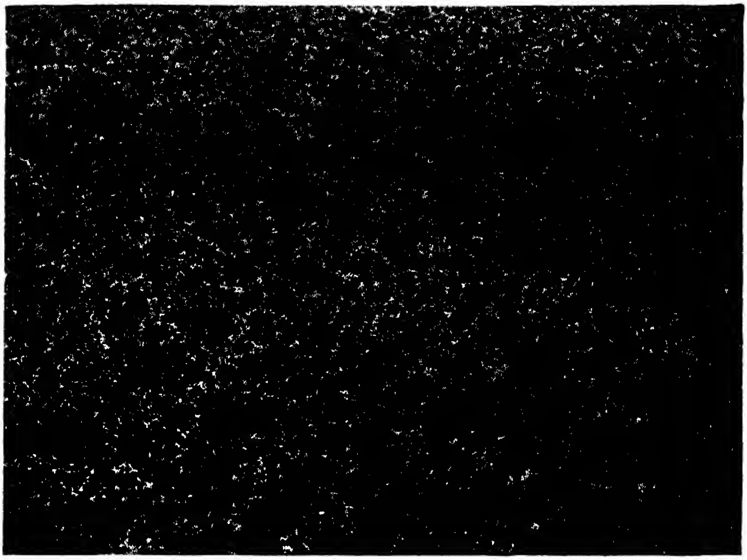

Fig. 3-XIII.—McQuaid-Ehn test. Fine-grained abnormal steel. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

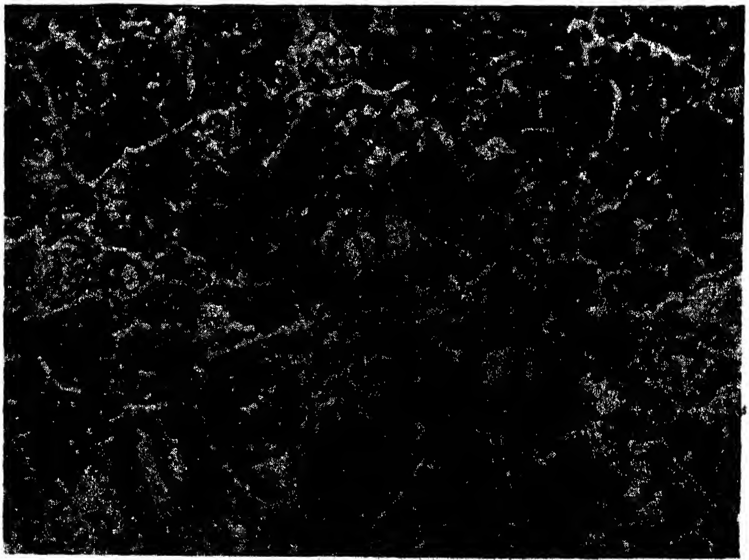

Fig. 4-XIII.—McQuaid-Ehn test. Fine-grained abnormal steel. × 1,000. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

grains outlined by proeutectoid ferrite, while the body of the grains will show the dark-etched fine pearlite.

In cases where the steel is only slightly hypoeutectoid with respect to carbon and in cases dealing with hypoeutectoid alloy steels, it is, in general, found satisfactory to cool the steel to about 1340°F. and hold the temperature there for a sufficient time before final cooling in air so as to allow ferrite to precipitate as a network, and then to quench in water. The subsequent quench transforms the remaining austenite grain to martensite. Proper etching reveals a network of ferrite or, in some instances, fine pearlite, surrounding the martensitic areas. Each martensitic area represents a single *transformed* austenite grain. This method is termed an *arrested quench.*

In cases where the steel is eutectoid in character, it is suggested that a specimen ⅝ to 1 in. in diameter be used, heated to the desired temperature, and quenched in water. In the quenching, one end of the specimen may be held out of the water. This procedure, which is termed a *differential quench*, produces a gradient structure effect. The unquenched end will show a zone of fine pearlite in which the grain size is not discernible, while the quenched end will be entirely martensitic. In between these zones will be one in which the cooling rate has been such that many nodules of fine pearlite will have had sufficient time to form in the boundaries of the grains while the grain body will be martensitic. On being etched, the fine pearlite will etch black and serve to outline the grains.

In hypereutectoid steels it is found satisfactory to use any of the procedures previously mentioned; the grains are outlined, in this case, by a fine carbide network. It is recommended, however, that, unless the carbon content is above 1.10 per cent, they be treated as eutectoid steels.

In cases where it is desirable to know the grain size in a steel that has already been quenched or quenched and tempered, the suggested procedure is to etch the suitably polished specimen with Vilella's martensite reagent. This reagent, which produces a contrast between the martensite of individual austenite grains, is made up as follows: Dissolve in 95 cc. of ethyl alcohol 1 g. of picric acid and 5 cc. of concentrated hydrochloric acid.

These methods of determining the austenitic grain size of steel without resorting to carburization present the following advantages:[1] (1) There is no danger of modifying the steel by the introduction of carbon and oxygen. (2) They may closely simulate any heat-treatment, especially as to the significant features, which are temperature, time at temperature, and (possibly) rate of heating. (3) A short time only is required to complete a determination of grain size.

Fracture Method.—In an earlier section, mention was made of the use of the Shepherd P-F (P = penetration of hardened structure; F = fracture ratings) test as a means of measuring hardenability. It has been found to be both easy and practical to combine the hardenability test with the grain-size evaluation by fracture as has been done by Shepherd. This test,[2] which has been carefully standardized, consists in hardening bars of definite section from four different heating temperatures, 1450, 1500, 1550 and 1600°F., fracturing the bars, and rating the resultant grain patterns with standards (sections hardened throughout) numbered serially from 1 to 10 (1, coarse and 10, very fine). Following this, one-half of the piece is polished and etched to show the depth of hardness, which is measured to the nearest $\frac{1}{64}$ in. The steel is then rated according to its fracture and the hardness depth when quenched from each of the four temperatures.

Vilella and Bain[3] have shown that there is a definite relationship between the austenitic grain size and the fracture grain size for plain carbon and low-alloy steels. They show that the grain sizes determined by fracture coincide substantially with the grain size determined microscopically and rated according to the A.S.T.M. grain-size chart.

This method of fracture rating is the most rapid procedure for the examination of steel for significant grain-size characteristics; if the size is uniform, the A.S.T.M. grain size may be reported at once; but if the fracture shows the probability of mixed grain size, it may be polished, etched, and rated microscopically by the standard chart.

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[1] Bain and Vilella, Austenitic Grain Size in Steel, A.S.M. "Metals Handbook," pp. 754–767, Cleveland, Ohio, 1939.

[2] Shepherd, B. F., The P-F Characteristic of Steel, *Trans. A.S.M.*, **22**, 979–1016 (1934).

[3] Vilella and Bain, Methods of Revealing Austenite Grain Size, *Metal Progress*, **30** (No. 3), 39–45 (1936).

GRAIN-SIZE CLASSIFICATION OF STEELS

After the representative specimen of the steel has been prepared in one of the previously outlined methods, the next step is to estimate the grain size. This may be accomplished by projecting the image of the specimen on a ground-glass screen at a given magnification and counting the number of grains in a definite known area. Another method is to compare the image with a series of standard photomicrographs of known grain count.

The A.S.T.M. has adopted a number chart based upon the number of grains per square inch of the image at a magnification of 100×. This standard chart, which is in wide use in the steel industry, covers all classes of steels. Grain-size numbers are used ranging from 1 to 8, in which each number covers a range in grain size. Table 1-XIII shows the grain-size numbers and the ranges covered by each.

TABLE 1-XIII*

Grain-size Number	Number of Grains per Square Inch at 100×
1	Up to 1½
2	1½– 3
3	3 – 6
4	6 –12
5	12 –24
6	24 –48
7	48 –96
8	96 and over

* With the present state of the art of steelmaking, the size ranges of Nos. 1 to 5 and Nos. 5 to 8 are suggested as standards for specifications.

This scheme of classification is applicable to all S.A.E. and allied structural steels.

AMERICAN SOCIETY FOR TESTING MATERIALS
STANDARD GRAIN-SIZE CHART FOR TENTATIVE CLASSIFICATION OF STEELS[1]

A.S.T.M. Designation: E 19-39T

This chart is issued under the fixed designation E 19; the final number indicates the year of original adoption as standard or, in the case of revision, the year of last revision.

Revised, 1939

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[1] Under the standardization procedure of the Society, this chart is under the jurisdiction of the A.S.T.M. Committee E-4 on Metallography.

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Scope.

1. This chart is intended to be used primarily for classification of S.A.E. and allied structural steels according to grain size which concerns only the size of the pearlite grains and is independent of the condition of the excess carbide.

2. It is recommended that grain-size studies be made only on specimens having prepared surfaces which are free from oxidation, decarburization, and influence of cold work.

Treatment of Specimen.

The specimen shall be carburized at 1700°F. (927°C.) for not less than eight hours in a compound which will produce a hypereutectoid zone. Any of the highly energized compounds on the market will be satisfactory for this purpose. The specimen must be cooled slowly enough to produce a pearlitic structure. This is usually accomplished by furnace cooling. After cooling, a microsection shall be prepared and the carburized zone examined at 100 diameters magnification and then compared with the grain-size chart shown on pages 306–313 inclusive.

Grain-size Classification.

3. (a) Sizes Nos. 1 to 5 may be considered "coarse-grained" steels and are generally characterized by complete carbide envelopes in the hypereutectoid zone.
   (b) Sizes Nos. 5 to 8 may be considered "fine-grained" steels and are generally characterized by incomplete carbide envelopes in the hypereutectoid zone.
   (c) Size No. 5 may be considered as either a "coarse-grained" or "fine-grained" steel depending on whether those relatively few grains of the hypereutectoid zone which are outside of the No. 5 range are mostly coarser or finer than No. 5.
   (d) The hypoeutectoid zone is included as an aid in determining grain size. It also gives an indication of hardening characteristics. In the hypoeutectoid zone, the condition of the ferrite will follow the same trend as the carbide in the hypereutectoid zone. Size No. 1 will usually show a maximum tendency for ferrite envelopes, while size No. 8 will usually show a maximum tendency for ferrite islands.

In earlier discussions of grain size it was emphasized that in referring to the size of grain the correct description should be in terms of three dimensions or volume; hence, the size in fractions of a cubic inch or cubic centimeter or, better, the number of

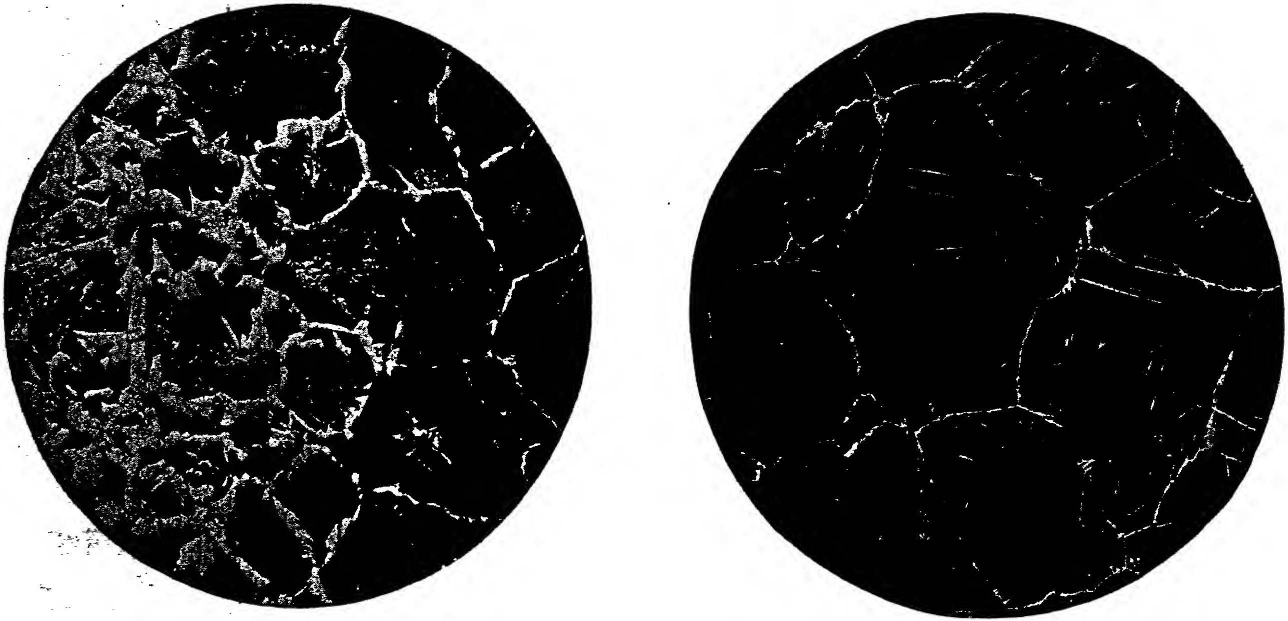

Fig. 5-XIII.—No. 1, up to 1½ grains per square inch.

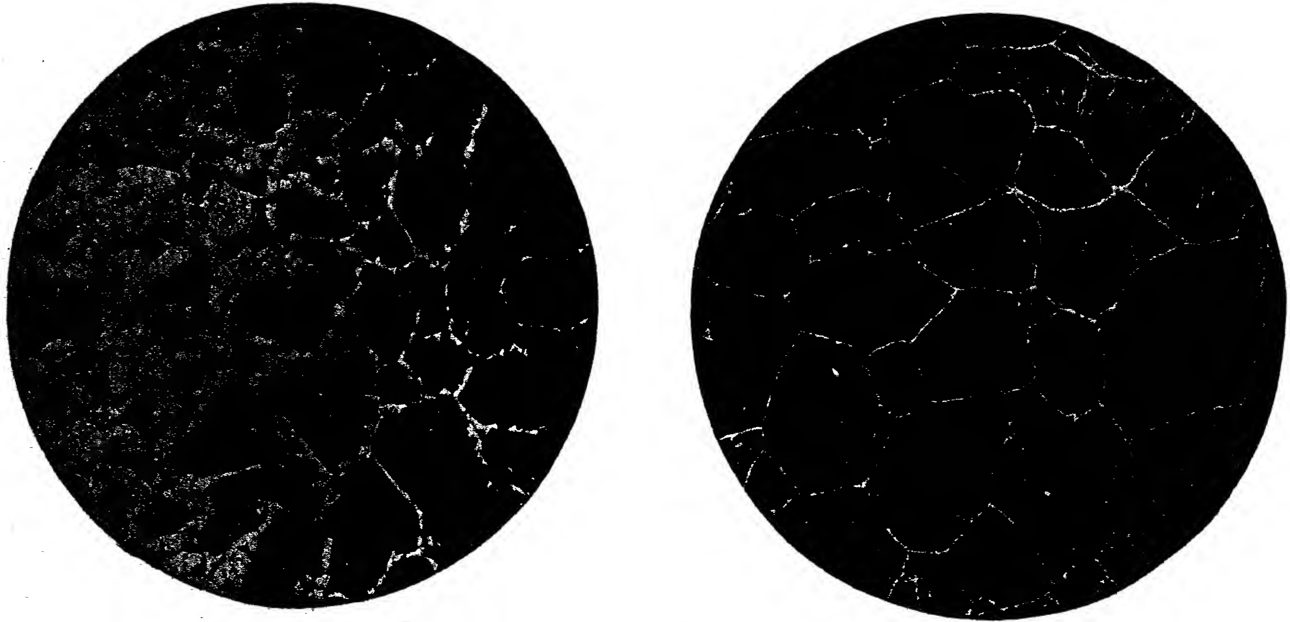

Fig. 6-XIII.—No. 2, 1½ to 3 grains per square inch.

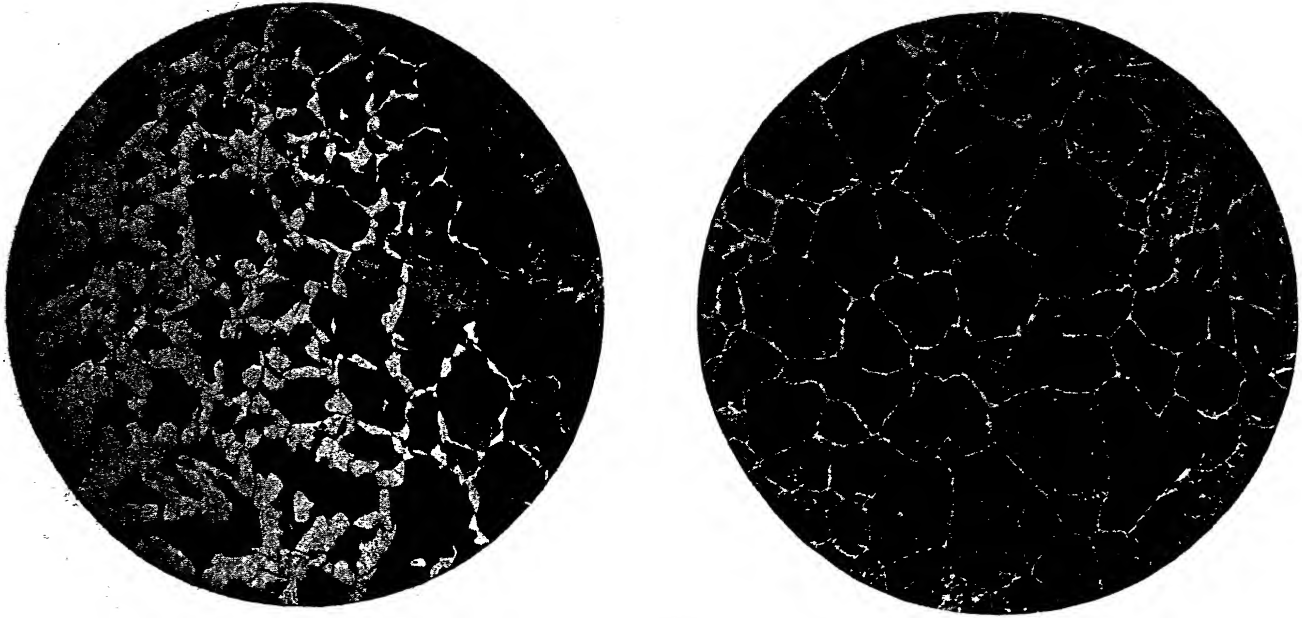

FIG. 7-XIII.—No. 3, 3 to 6 grains per square inch.

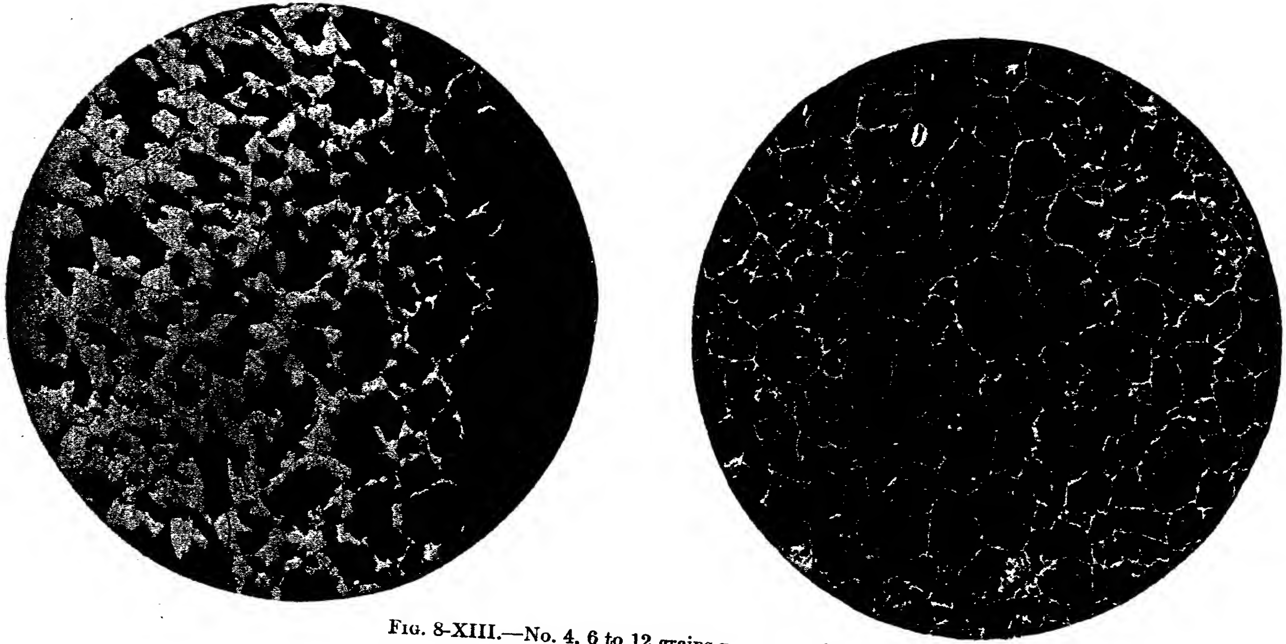

Fig. 8-XIII.—No. 4, 6 to 12 grains per square inch.

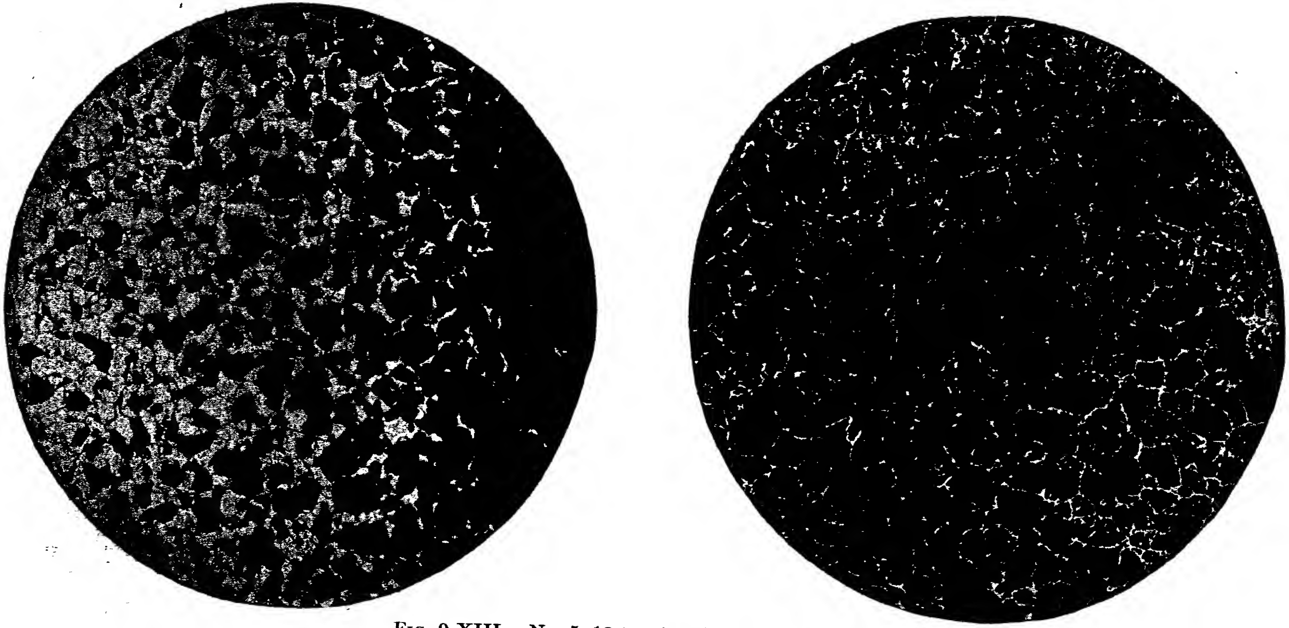

FIG. 9-XIII.—No. 5, 12 to 24 grains per square inch.

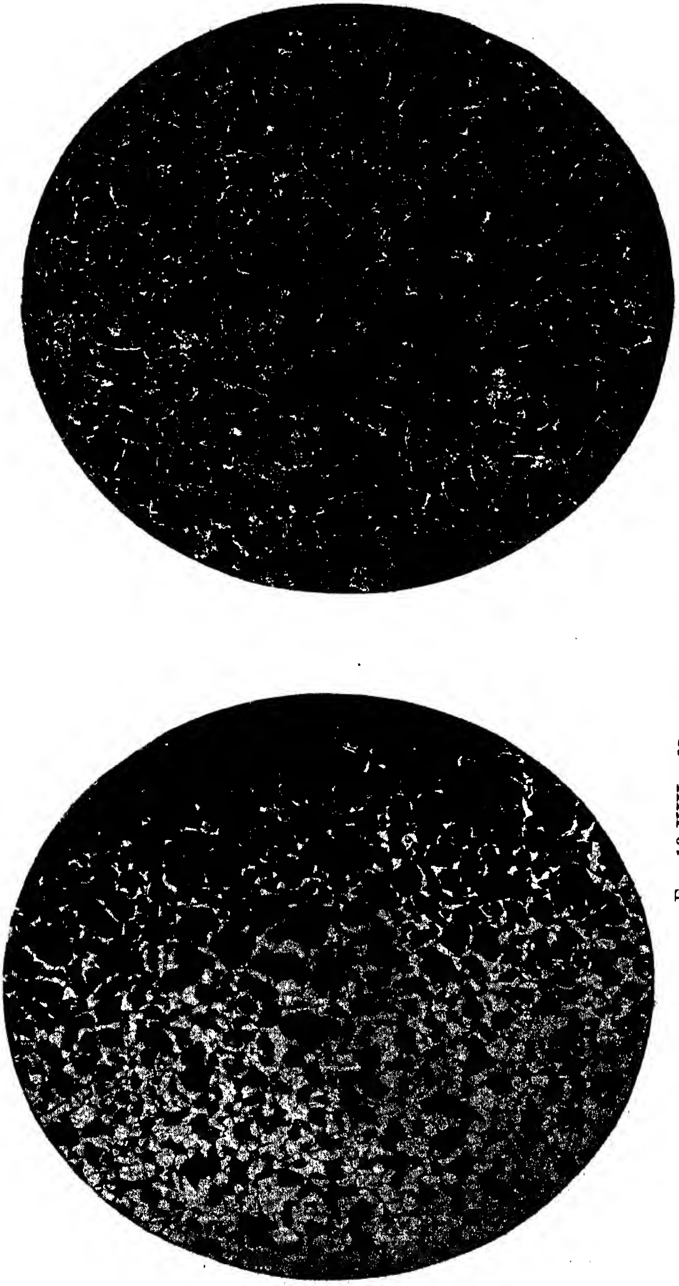

FIG. 10-XIII.—No. 6, 24 to 48 grains per square inch.

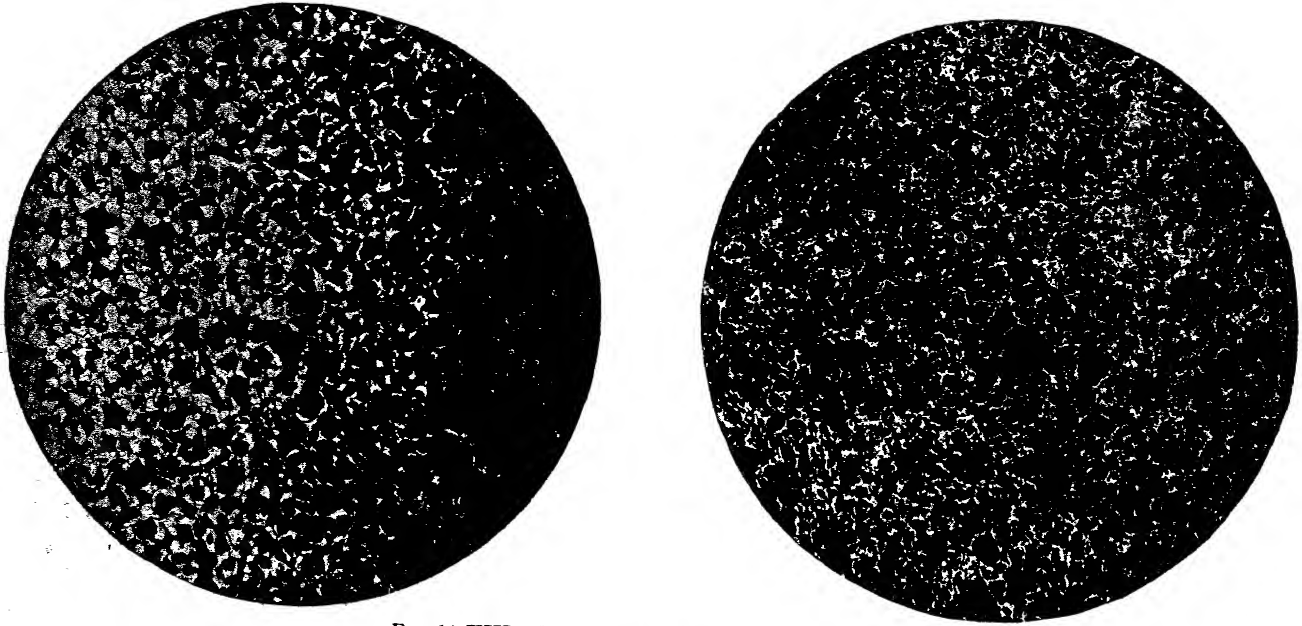

Fig. 11-XIII.—No. 7, 48 to 96 grains per square inch.

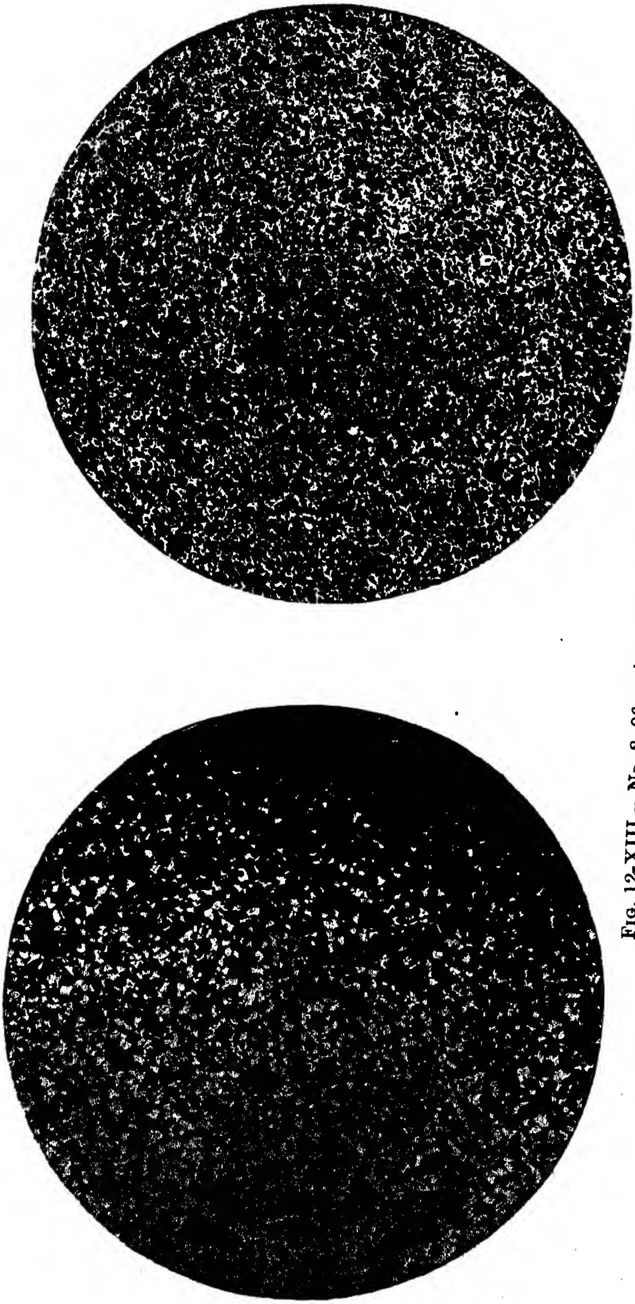

FIG. 12-XIII.—No. 8, 96 grains and more per square inch.

grains per cubic inch or per cubic centimeter. It has been found, however, that a plane-section examination is entirely satisfactory.

In making a grain-size count on a plane section at 100 × it should be obvious that this plane will cut the mass of the grains at their maximum cross section, while the remaining ones will show only a small cross section where the plane has, for example, cut only their corners. When the examination shows a number of small grains regularly distributed with larger grains, this is merely the result of the geometry of the section. When, however, the small grains are grouped in a cluster, this indicates a "duplex," or "mixed" grain. Either this condition is the result of inhomogeneity of the steel or it indicates that the coarsening process was caught in an intermediate stage, so that equilibrium for that temperature had not been reached, with the result that the absorption of the small grains by the large grains was irregular. Figure 13-XIII illustrates such a condition.

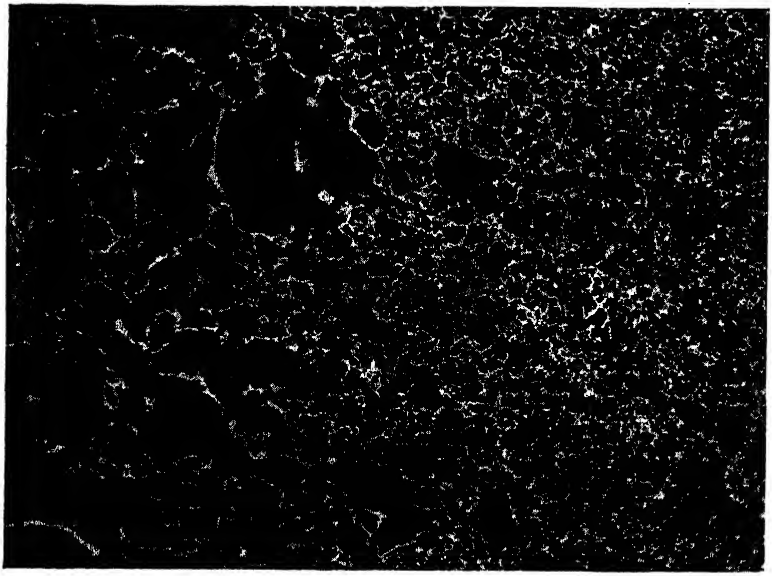

FIG. 13-XIII.—Photomicrograph of a duplex McQuaid-Ehn grain size test sample. × 100. Nital etched. (*Courtesy of Bethlehem Steel Company.*)

GRAIN-SIZE CONTROL

A study of grain-size control of open-hearth steel by many investigators has shown that with ordinarily good furnace practice a positive production of coarse- or fine-grained carbon steels can be attained by a controlled ladle addition of aluminum and a controlled reactive oxygen content, which in turn is largely dependent upon the carbon, manganese, and silicon content of the steel. Thus,[1] for a range of carbon steels of between 0.15 and 0.50 per cent, manganese 0.50 per cent and above, and silicon content of about 0.15, a fine-grained steel can invariably be obtained, for a given furnace practice, by the addition of about 1 lb. of aluminum per ton in the ladle. When the silicon content was 0.10 per cent or less, a coarse-grained steel was obtained by omitting the aluminum or by using considerably less aluminum than was used in the first practice, say, ½ lb. per ton in the ladle. "Particular emphasis, it has been found, must be placed on deoxidation control since it is now known that the most important factor in controlling the effect of the aluminum used is in the degree of oxidation and the type of oxides present when the deoxidizers are added." It has also been found that different results are obtained according to whether the aluminum is added to the furnace, the ladle, or the mold. In general, smaller quantities of aluminum are necessary to yield fine grains in the mold: to produce the same results, larger quantities are needed when it is added to the ladle; and much larger quantities must be used when it is added to the furnace.

Deoxidation, as we know, is based on the formation, in the molten steel, of insoluble oxides by the addition of certain elements that have a greater affinity for oxygen than does iron. After the iron has been purified—as far as the undesirable impurities are concerned—it becomes necessary to eliminate the FeO and the gases that are soluble in the molten bath. This is accomplished by the addition of certain elements that will react with FeO and the gases in the metal bath and form nonmetallics, which must then be eliminated. The nonmetallics formed by the deoxidizers will be distributed throughout the bath as a dispersion. According to Herty,[2] the type of dispersion thus

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[1] Epstein, Nead, and Washburn, Grain Size Control of Open Hearth Carbon Steels, *Trans. A.S.M.*, **22**, 942–978 (1934).

[2] Herty, McBride, and Hough, The Effect of Deoxidation on Grain Size and Grain Growth In Plain Carbon Steels, *Cooperative Bulletin* 65, Carnegie Institute of Technology and Mining and Metallurgical Advisory Boards, Pittsburgh, 1934.

formed is dependent upon the iron-oxide content of the metal before deoxidation, the amounts of the various deoxidizers used, the relative times of their additions, and the degree of fluxing of the particles with each other and with residual dissolved iron oxide following the deoxidation process. Since fluxing of the impurities is of considerable importance, the temperature of the metal must be added, as another variable. There may also be present certain dispersions, owing to their presence in the charge.

The type and quantity of these deoxidation products remaining in the metal will determine its physical characteristics and its adaptability. Therefore, deoxidation besides eliminating FeO has its effect upon the character of the metal. By the addition of the usual deoxidizers, such as manganese, silicon, or aluminum, their respective oxides are formed, which are insoluble in the liquid steel. SiO_2, which is acid, FeO and MnO, which are basic, will combine to form slags of low melting points, which will tend to coalesce to larger particles and will rise from the metal. Al_2O_3 particles, because of their high melting point, will not coalesce readily and, owing to the fact that they are nearly neutral with respect to acidity or basicity, there is little tendency for them to combine with FeO and MnO or both to form a slag of low melting point and to be fluxed off. Elimination of Al_2O_3 is very slow, much slower than either SiO_2 or MnO, and this makes the use of aluminum alone as a deoxidizer undesirable.

Aluminum, although it is a much more powerful deoxidizer than either silicon or manganese, has been looked upon for many years as extremely undesirable because of the amount of residual oxide left in the steel. However, in the manufacture of fine-grained steel it has become of real value and importance. The usual practice is to deoxidize the metal first by the addition of silicon and manganese in the furnace, and to follow this by definite aluminum addition in the ladle or mold. (The amount of stronger deoxidizer needed to control the grain size is not much greater than that frequently used in killed steels.) Aluminum thus added carries the deoxidation further and leads to the formation of numerous finely dispersed particles of Al_2O_3. These particles presumably act as nuclei for the formation of the austenite grains and may offer obstruction to grain growth. Large particles in the metal seem to have little effect on grain size, and it is found that only after the deoxidation has been carried to the point where submicroscopic particles are formed does the grain-size control become apparent.

With these facts in mind, let us make a brief résumé of the conditions. Starting with the metal entering the mold, we have found that it contains iron oxide in solution, other oxides in solution, oxides in suspension, and certain alloying elements to meet the chemical specifications. The oxide suspensions will vary in size from those that are colloidal in nature to those that are large enough to float out of the steel. After teeming, solidification starts and some of the soluble material becomes insoluble, through a change in solubility that may be due to the lowering of the temperature, or through a chemical action, or through both. During this time, some of the particles may either agglomerate or coalesce and form large inclusions, which may be observed under the microscope if they have not had sufficient time to rise through the metal. The remaining particles in suspension or dispersion may act as nuclei for the formation of more grains than would normally be formed if the metal were free of such dispersions and would tend to prevent grain growth, once the austenite grains had been formed.

It is pertinent at this point to emphasize that the theory here presented to account for the austenitic grain-growth characteristics—*i.e.*, the presence of numerous submicroscopic particles of a refractory nature[1]—is just one of the explanations that have been advanced. It is generally agreed that the facts we have just observed are a direct result of deoxidation with aluminum, but the substance and the mechanism by which the grain-growth inhibition is accomplished are still under discussion. A complete discussion of the merits of the various theories can be obtained from the reference in the footnote.[1]

On the basis of the previous discussion and our assumption regarding particle size, presumably when such deoxidizers are used as tend to form large particles, we should get coarse-grained steels. Herty[2] in his investigations has shown that silicon killed

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[1] Ward and Dorn, Grain Size of Steel, *Metals and Alloys*, **10**, 250–253 (August, 1939).

[2] Bain, E. C., Factors Affecting the Inherent Hardenability of Steel, *Trans. A.S.S.T.* **20**, 385–428 (1932); Herty, McBride, and Hough, The Effect of Deoxidation on Grain Size and Grain Growth in Plain Carbon Steels, *Cooperative Bulletin* 65, Carnegie Institute of Technology and Mining and Metallurgical Advisory Boards, Pittsburgh, 1934.

steels of from 0.08 to 0.10 silicon were coarse-grained, while those having a silicon content of from 0.15 to 0.20 were consistently finer grained. He postulates that this difference is due to the fact that as the silicon content increases there is less opportunity for the FeO and MnO to flux the silica particles. Steels with low silicon content contain almost entirely large inclusions, while as the silicon content was increased the silica particles were present in smaller form.

Aluminum-treated steels may also be coarse-grained. It has been found that if the equivalent amount of aluminum, instead of being added to the ladle (thereby yielding a fine-grained steel), is added to the furnace, the resultant steel will be similar to that of a silicon-killed steel. Under these conditions, the alumina particles formed have been fluxed and have lost their ability to act as nuclei for crystallization.

It should be noted that, if the deoxidizing element is added in too great an excess, the resulting iron alloy will have different grain characteristics from those that would be found in iron if the amount of deoxidizing element were low. For example, "3.5 per cent silicon, 1.5 per cent manganese, and 0.50 per cent aluminum steels are all generally coarse-grained. These quantities as you can see are far in excess of those used for deoxidation alone."

FORMATION AND GROWTH OF AUSTENITE GRAINS

When a piece of steel of either carbon or low-alloy analysis is heated above the critical point, the ferrite and carbide will react with one another to form austenite. This reaction starts at nuclei located in ferrite-cementite interfaces. Such small austenite grains continue to grow by absorbing the adjacent ferrite and cementite until the transformation is complete. As a general case, some of these grains will grow more rapidly than others, so that when a large grain encounters a small one it will be absorbed. This particular action is found to occur frequently in some steels and rarely in others.

If a piece of steel of coarse-grained character is heated through the critical it will, after recrystallization, start to coarsen and will continue to coarsen as the temperature increases up to a point beyond which there will be very slight change. The grain size after recrystallization with silicon-killed plain carbon steels[1]

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[1] Herty, McBride, and Hough, The Effect of Deoxidation on Grain Size

(coarse grain) is dependent upon the rate of heating through the critical range. With increasing rate the initial austenitic grain size becomes larger. These grains will continue to grow until the steel has been at the heating temperature for about 20 min., after which the grain size remains fairly constant. In plain carbon steels, that have had no aluminum or equivalent alloying additions, the grains will acquire a No. 1 or No. 2 A.S.T.M. size over a wide range of temperatures.

A steel with fine-grain characteristics (aluminum treated), on the other hand, has very different growth characteristics.[1] The rate of heating through the critical range has very little effect on the initial size of the austenite grains. All heating rates will produce small grains of No. 5 to No. 8 A.S.T.M. size.

The austenitic grains of aluminum-killed steels are found to maintain this small initial size practically indefinitely at temperatures below what is known as the coarsening temperature. If a steel of these characteristics is heated to a series of increasing temperatures and the respective austenitic grain sizes are studied, a temperature will be reached where grain growth will occur. This temperature or range of temperatures is known as the coarsening temperature or the coarsening range for the steel. Above this temperature large grains are produced, which, in general become larger than would be produced under the same temperature conditions with similar unkilled steels. The composite photomicrographs, (Fig. 14-XIII)[2] illustrate the difference in grain-coarsening characteristics between coarse- and fine-grained steels that were normalized at the temperatures shown for 1 hr. The austenitic grains in the Fig. 14-XIII are outlined by ferrite.

The effect of time and temperature on grain growth as indicated by air cooling (normalizing) is illustrated in Table 2-XIII by four steels taken from the work of Herty, McBride, and Hough.[1]

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and Grain Growth in Plain Carbon Steels, *Cooperative Bulletin* 65, Carnegie Institute of Technology and Mining and Metallurgical Advisory Boards, Pittsburgh, 1934.

[1] Herty, McBride, and Hough, The Effect of Deoxidation on Grain Size and Grain Growth in Plain Carbon Steels, *Cooperative Bulletin* 65, Carnegie Institute of Technology and Mining and Metallurgical Advisory Boards, Pittsburgh, 1934.

[2] Herty, McBride, and Hollenback, Which Grain Size? *Trans. A.S.M.*, **25**, 297–314, (1937).

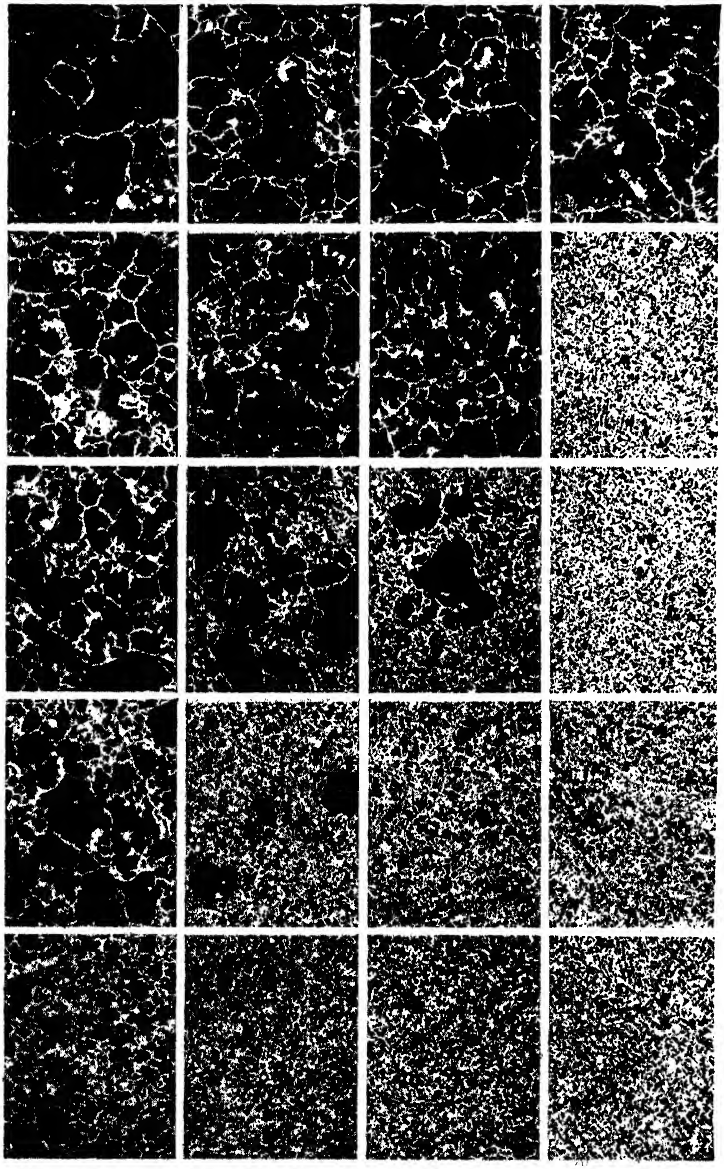

FIG. 14-XIII.—Effect of temperature on the grain size of normalized steel. × 100. Reduced 43 per cent in reproducing for printing. (*Courtesy of American Society for Metals.*)

TABLE 2-XIII

	C	Mn	P	S	Si	Metallic Al
Silicon killed (J)........	0.39	0.73	0.016	0.031	0.073	
Silicon killed (Y)........	0.41	0.62	0.021	0.034	0.191	
Aluminum killed (A)...	0.48	0.74	0.022	0.030	0.205	0.020
Aluminum killed (Z)....	0.41	0.58	0.018	0.031	0.183	0.016

The grain growth characteristics of these steels for given times at temperatures of 950 and 1000°C. (1742 and 1832°F.) are shown in Table 3-XIII.

TABLE 3-XIII.—EFFECT OF TIME ON GRAIN SIZE AT 1742°F. (950°C.), GRAIN SIZE EXPRESSED AS GRAINS P.S.I. AT 100 ×

Steel	Time at temperature (min.)					
	0.0	15	30	45	60	90
Aluminum killed.......	7.7	6.8	6.3	6.1	5.9	5.5
Aluminum killed.......	151.0	151.0	145.0	136.0	133.0	129.0
Silicon killed..........	10.5	5.1	4.2	3.9	3.9	3.8
Silicon killed..........	13.5	7.4	5.1	5.1	4.3	4.2

at 1832°F. (1000°C.)

Steel	Time at temperature (min.)			
	0.0	20	40	60
Aluminum killed..............................	7.4	6.2	5.8	4.6
Aluminum killed..............................	98.0	2.8	2.0	2.0
Silicon killed..............................	5.7	4.1	3.8	3.3
Silicon killed..............................	6.1	4.4	3.5	2.9

In general, it was found that the grain growth at 950 and 1000°C. was essentially complete in about 20 min. at temperature, for all the steels. At 950°C., as can be seen, only the steel *A*, which had been treated with an excess of aluminum, remained fine-grained, even after heating for 90 min. at temperature. At a higher temperature, this steel along with the others became coarse-grained.

The effect of grain growth and normalizing temperature is shown in Fig. 15-XIII. The same four steels as in the above analysis are used to illustrate this effect. Suitable specimens were placed in the furnace, which was at the normalizing heat, and held at that temperature for 1 hr., starting from the time the specimens reached the furnace temperature, were air cooled and then examined for grain size. The silicon-killed steels, as can be seen, gradually coarsen as the heating temperature is raised. After reaching a temperature of 1742°F. (950°C.) there is very little grain growth. The aluminum-treated steels show abrupt changes in grain size at about 1697 to 1787°F. (925 to 975°C.). After complete grain growth at the higher temperatures, these

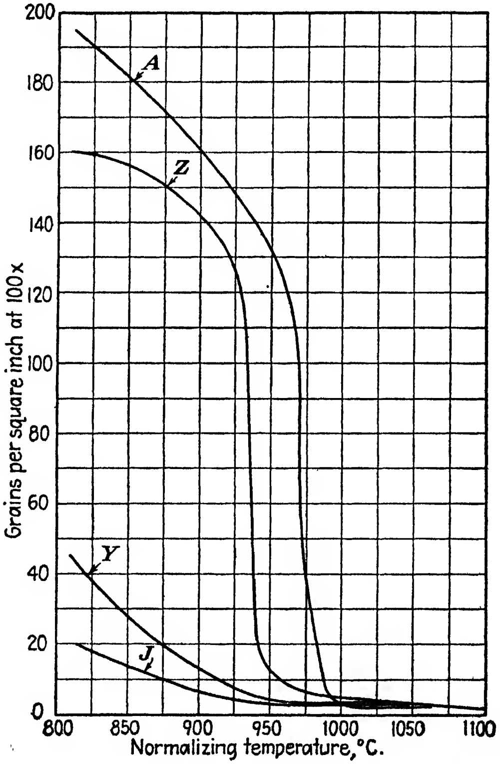

FIG. 15-XIII.—Grain size of samples held at various temperatures for one hour, air cooled. (*After C. H. Herty, Jr.*)

steels will have, on the average, coarser grain than the originally medium-grained steels.

It is interesting to note that the aluminum-killed steel (*Z*), according to Table 3-XIII, is coarse-grained at 950°C. This illustrates an important point concerning aluminum-killed steels. It has been shown that the trends illustrated are typical but that they all differ with respect to the temperature at which they coarsen.

From the previous discussion it should be apparent that for heat-treating operations the important factor is the coarsening temperature of the steel. Below this temperature the steel will be fine-grained, regardless of the holding time, while above the temperature it will have characteristics typical of a coarse-grained steel. Thus, if the steel has a coarsening temperature of 1600°F. and it is desirable that it be coarse for one operation and fine for a second, it can be coarsened by heating to 1650°F. for the first and then re-treated to 1550°F. for the second. The success of such treatment is dependent upon the determination of the true coarsening temperature of the steel. This emphasizes the point that such a test as the McQuaid-Ehn will not be satisfactory unless the steel is to be treated under the conditions of the test.

There are other factors besides time and temperature that alter the austenitic grain-growth characteristics of steel. Cold work, in general, is found to produce finer austenitic-grained steels.[1,3] Hot work may reduce the grain-coarsening temperature several hundred degrees.[1,3] Heat-treatment[2,3] is also a factor. In fine-grained steels in which the coarsening temperature is high, the coarsening temperature will be lowered 75 to 100° by normalizing in the customary range of 1600 to 1700°F. The amount of lowering varies with different steels and with temperature. Increasingly higher normalizing temperatures cause less lowering, so that at about 1900 to 2000° there may be none at all. This response, however, varies, so that some will have theirs lowered while others will have theirs raised. With other rates of

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[1] Grossmann, M. A., On Grain Size and Grain Growth, *Trans. A.S.T.*, **21**, 1079–1111 (1933).

[2] "U.S.S. Carilloy Steels," pp. 93–112, Carnegie-Illinois Steel Corp., Pittsburgh, 1938.

[3] Grossmann, M. A., Grain Size in Metals, with Special Reference to Grain Growth in Austenite, *Trans. A.S.M.*, **22**, pp. 861–878 (1934).

TABLE 4-XIII.—EFFECT OF AUSTENITIC GRAIN SIZE ON THE PROPERTIES OF STEEL

Grain	Property	Effect of finer austenitic grains
In the austenitic condition	Plastic forming	Poorer forgeability and greater resistance to slip
	Rate of carburizing and decarburizing	Slower
	Rate of decomposition	More rapid
Annealed or normalized...	Machinability	Poorer, owing to poorer chip-forming properties
	Machined surface	Better surface finish
	Cold working	Less embrittled, possible to draw deeper and roll more severely
	Elastic ratio	Higher
	Yield point	Slightly lower
	Ultimate strength	Slightly lower
	Elongation	Greater
	Reduction in area	Greater
	Impact resistance	Higher
	Hardness	Lower
	Creep resistance	Better below equicohesive temperature Poorer above equicohesive temperature
	Normality	Greater tendency toward abnormality
	Difference between annealed and normalized structures	Less
Hardened or hardened and tempered.	Maximum hardness	Independent of grain size
	Hardenability	Much shallower hardening
	Impact resistance	Much greater
	Quench cracks	Seldom exhibits quench cracks
	Grinding cracks	Seldom shows grinding cracks
	Untransformed austenite	Less austenite retained
	Soft spots	More susceptible to soft spots
	Distortion	Less susceptible
	Internal stress	Lower

cooling, such as annealing or quenching in oil or water, the coarsening temperature is lowered, though not in exactly the same manner or to the same degree.

INFLUENCE OF AUSTENITIC GRAIN SIZE ON THE PROPERTIES OF STEELS

Since it is now well recognized that both coarse and fine austenitic grains have definite advantages for particular applications, treatments have been developed to produce the grain size that responds most favorably to a given operation. Table 4-XIII,[1] which shows the effect of austenitic grain size on various properties of steels, may be used as a guide in setting up the proper treatment. A complete bibliography covering all the properties noted in this table may be obtained from the review by Ward and Dorn.[1]

Hardenability.—The most important feature of grain size undoubtedly lies in its influence on hardenability, since it is found that the properties of quenched steels are more directly related to the austenitic grain size than are the properties of annealed or normalized steels. According to the previous discussion, hardenability was found to be a measure of the maximum hardness attainable and the penetration or depth of hardening. Thus, if a round is quenched from a temperature above the upper critical in a medium that will cool the outside layer at a rate exceeding the critical cooling rate, this part will transform to martensite and have a maximum hardness consistent with the chemical analysis of the steel. The inside layers, however, will cool at rates slower than the critical, depending upon the section size, and will not develop maximum hardness.

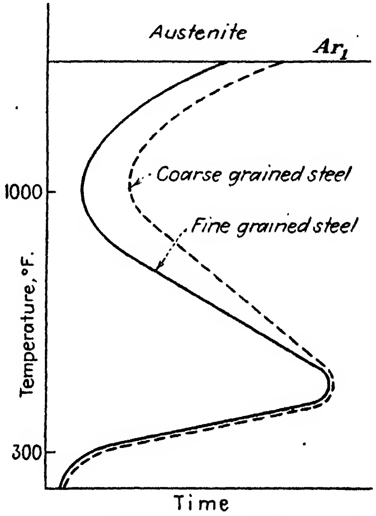

FIG. 16-XIII.—Comparative reaction rate curves for fine and coarse-grained steel of the same chemical analysis. (*After A. A. Bates.*)

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[1] WARD and DORN, Grain Size in Steel, *Metals and Alloys*, **10**, 74–81, 212–218 (1939).

It has been found[1] that the rate at which austenite will decompose is a direct function of the grain surface[2] for any given cooling rate, the larger the grain surface—*i.e.*, the smaller the grain size—the more rapid the decomposition. Therefore, the critical cooling rate is greater for fine-grained than for coarse-grained steels. From the standpoint of heat-treatment, this indicates that the difference between a fine- and coarse-grained steel is primarily one of reaction rates. This difference is graphically illustrated in Fig. 16-XIII. It shows that at the high temperature transformation area (1000°F.) the fine-grained steel will transform at greater speeds than will the coarse-grained. For full hardening, then, it will be necessary to cool a fine-grained steel at a greater rate than is required for a coarse-grained steel.

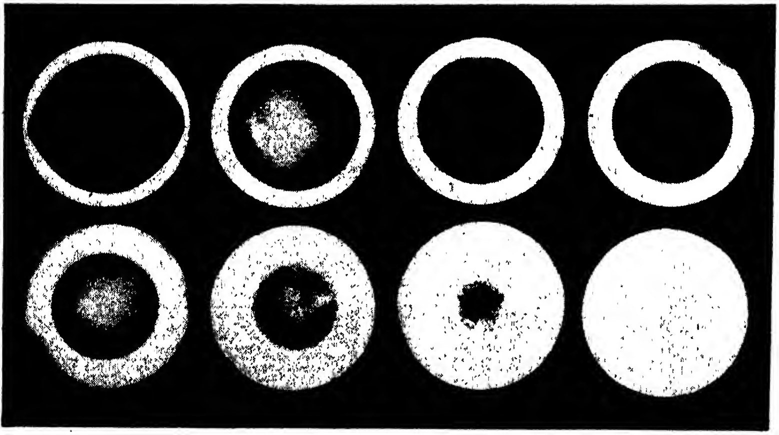

Fig. 17-XIII.—Sections of hardened one-inch rounds of 1.0 per cent carbon steels of different hardenability at time of quenching. (*Courtesy of the Research Laboratory, United States Steel Corporation.*)

The variation of the depth or penetration of hardening is illustrated in Fig. 17-XIII, which shows a series of 1-in. rounds of 1.00 per cent carbon steel with varying austenitic grain size, all quenched under identical conditions. The steels vary from fine-grained at the upper left to coarse-grained at the lower right. This shows that the bar of fine-grained steel of this diameter

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[1] Herty, McBride, and Hollenback, Which Grain Size? *Trans. A.S.M.*, **25**, 297–314, (1937).

[2] Assume that the grains are spheres.

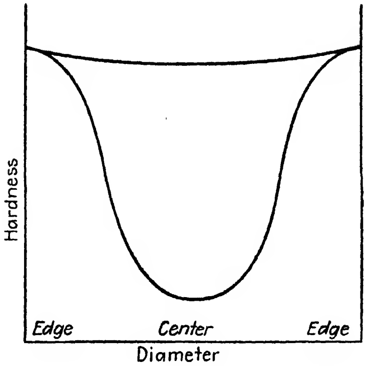

FIG. 18-XIII.—Curves illustrating the variation of hardenability and grain size. (Upper) Deep hardening coarse-grained steel. (Lower) Shallow hardening fine-grained steel.

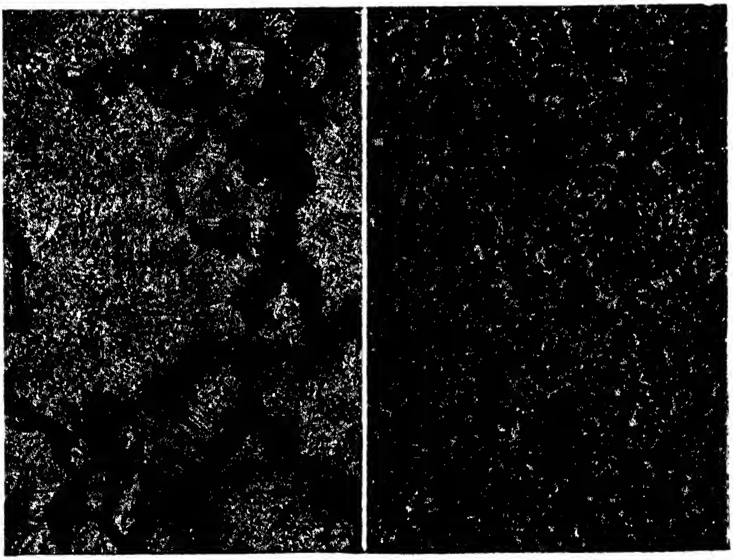

FIG. 19-XIII.—Grain size characteristics in unkilled and aluminum killed 0.39 per cent carbon steels as identically quenched from 1600°F. (870°C.) (Left) Coarse-grained unkilled. (Right) Fine-grained aluminum killed. Microstructure in the interior of quenched 1-in. rounds. × 100. (*Courtesy of American Society for Metals.*)

under the quenching conditions was not cooled fast enough to harden fully to its center. On the other hand, the coarse-grained bar under the same conditions was completely hardened throughout its cross section. In general, types of steel that have fine austenite grains are so-called *shallow hardening*, while those having coarse austenite grains are *deep hardening*. This effect is further illustrated in Fig. 18-XIII, in which the upper curve schematically represents a deep-hardening, coarse-grained steel and the lower curve a shallow-hardening fine-grained steel. Contrasting microstructures developed by quenching a 0.40 per cent carbon steel of fine- and coarse-grained character is shown in Fig. 19-XIII.[1]

The methods devised for the determination of hardenability have been discussed in Chap. X.

SPECIFICATION BY CLASSES FOR CARBON STEELS[2]

Plain carbon steels, not heat-treated (such as S.A.E. 1010, 1020, etc.), should have a coarse grain for maximum machinability and best cold-heading properties. A fine grain is better when the steel is to be subjected to shock load or occasional severe overstress.

Plain carbon steels, heat-treated, uncarburized parts (S.A.E. 1025, 1040, 1050, etc.) should be generally of a fine-grained, shallow-hardening type. When they are oil hardened to high Brinell, use a coarse-grained type. Use a coarse-grained type in very heavy parts such as truck front axles; medium-grained in passenger-car front axles and on parts with heavy and light sections—the heavier sections will harden satisfactorily without cracking the light sections.

In plain carbon, case-hardening steels (S.A.E. 1010, 1020, X1315), a fine grain is used for minimum distortion; a coarse grain for machining and deep hardening. The general recommendation is that they should be as fine as possible, consistent with satisfactory surface hardness, so as to avoid corner cracking of parts hardened very deeply and grinding cracks. They are used in camshafts, small gears, differential parts, brake parts, and washers.

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[1] DAVENPORT and BAIN, General Relations between Grain Size and Hardenability and the Normality of Steels, *Trans. A.S.M.*, **22**, 879–925 (1934).

[2] MCQUAID, H. W., Effect of McQuaid-Ehn Grain Size on Hardness and Toughness of Automobile Steels, *Trans. A.S.M.*, **22**, 1017–1037 (1934).

A plain carbon, high carbon type of steel (S.A.E. 1060, 1095, etc.) is useful for flat springs, valve springs, and coil springs. A fine grain is best to avoid stress concentration. Bumper-bar stock may be of medium to coarse grain size, for oil hardening.

Suggested Questions for Study and Class Discussion

1. What tests might be used for determining the grain size of steels? What limitations must be made in regard to these tests?

2. Discuss the steel practice that might be used in producing a heat of fine-grained steel (6 to 8). How is it possible for the melter to produce other ranges of grain size?

3. Will hot rolling change this grain-size range? Discuss the effect of finishing temperatures.

4. In normalizing (air cooling), which is the more important, the effect of time or of temperature? Why?

5. Why is most of our interest, in regard to grain size limited to steels containing above about 0.15 per cent of carbon?

6. In processing steel balls for use in ball mills, a certain concern forges these balls at approximately 2300°F. and quenches them directly into oil.

a. What grain size should be specified for such steel? Microscopic examination of the cross section of the balls showed very deep hardening. One heat of steel when processed showed very shallow hardening. The balls were, thus, unsatisfactory, because of their tendency toward chipping or spalling.

b. What might have gone wrong with this particular heat of steel?

c. If this is a grain-size problem, would you use the standard McQuaid-Ehn test in your investigation?

7. In the discussion of patenting high carbon wire, mention was made of the fact that grain size was an important factor. Why was it desirable to use a coarse-grained steel?

8. In general, how does this factor of controlled grain size affect annealed, normalized, and quenched-and-drawn products?

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CHAPTER XIV

THE HEAT-TREATMENT OF STEEL CASTINGS

During the early history of the steel-casting business castings were made, shipped, and used in service without receiving any heat-treatment whatsoever; in other words, they were used in the "as-cast" condition. With increased service requirements, some of the castings were annealed—at least, those that were required to withstand severe service. In recent years the requirements have become such that castings are now being supplied in the full-annealed, the normalized, and the quenched and tempered condition. By suitable drawing and tempering, following the quench, a wide range of properties is ensured, which can be regulated at will to meet the necessary requirements.

CLASSIFICATION OF STEEL CASTINGS

Steel castings are divided into two main classes, carbon and alloy. Each of these divisions may be further subdivided according to its composition.

Carbon Cast Steels.—The carbon cast steels are divided into the following classes:[1]

1. The medium carbon cast steels, which constitute the bulk of the steel-foundry output, range from 0.20 to 0.50 per cent carbon, 0.50 to 1.00 per cent manganese, 0.20 to 0.75 per cent silicon, 0.05 per cent maximum phosphorus, and 0.06 per cent maximum sulfur.
2. The low carbon cast steels, which are generally employed in special applications, range from 0.20 per cent carbon and below, with the above specifications on the other elements.
3. The high carbon cast steels, which are employed in applications where considerable strength, resistance to abrasion and toughness are required, range over 0.50 per cent carbon, with the above specifications on the other elements.

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[1] A.S.M. "Metals Handbook," p. 959, Cleveland, Ohio, 1939.

Owing to the large number of element combinations that it is possible to use in the production of alloy cast steels, and the enormous range in physical properties obtainable through the composition and heat-treatment, only a general discussion will be presented of that class. The following practices will, therefore, apply to carbon and alloy steels, where the alloy steels refer to the commercial castings used for ordinary construction purposes and not for special uses. A number of the special casting alloys will be discussed in Chap. XX.

HEAT-TREATMENT

After the castings have been poured, allowed to cool to a temperature below the critical range, and cleaned, they are ready to be heat-treated.

The heat-treatment of steel castings is essentially the same as that used for rolled or wrought products. The important difference is that steel castings in the as-cast condition have a very coarse and frequently a dendritic structure. Steel in the as-cast state for rolled and wrought products will show the same type of structure, but in these cases it is broken up by subsequent rolling and forging operations. To bring about the proper refinement in castings and develop structures and physical properties similar to those of wrought material of similar composition, the castings are heated to a temperature well above the upper critical-temperature range and are cooled at various rates. The treatment given to a casting will vary with respect to its industrial application.

Single Full Anneal or Single Normalize.—The recommended practice according to the A.S.M. "Metals Handbook" is given in Table 1-XIV.

TABLE 1-XIV

Carbon range	Temperature, degrees F.	Normalizing, cool in	Annealing, cool in
0.15–0.40	1600–1700	Still air	Furnace or other medium for cooling slowly*
0.40–0.60	1550–1600	Still air	Furnace or other medium for cooling slowly

* The castings are cooling slowly until the temperature has dropped to 800°F. or below.

Both of these treatments have as their purpose the breaking up or diffusing of the dendritic segregation as far as is possible, so as to form a homogeneous structure and an accompanying uniformity of grain size. In particular, full annealing will effect the removal of stresses, improve machinability and ductility. Normalizing will produce a finer grain structure as a result of the more rapid rate of cooling. It will, thus, increase the physical properties over those attainable on the same steel on full annealing. This increase in properties is so marked in many instances that it has received widespread adoption in the treatment of both small and medium-sized castings.

In the higher carbon steels, normalizing induces stresses and brittleness, which usually necessitates a reheat at temperatures from 1000 to 1300°F., in order to increase the ductility and reduce the stress.

The heating temperatures recommended for the single treatments are compromise temperatures, which are approximately 200°F. above the critical range. The compromise was developed in order to bring about a balance between time and temperature for proper diffusion and grain growth. A temperature just above the critical would produce recrystallization but the rate of diffusion would be extremely slow, while with increase in temperature the rate of diffusion will increase but, unless the steel is resistant to grain growth, the resultant grain size will also increase rapidly. Hence, the recommended temperatures are compromises that will give diffusion in a reasonable time without excessive grain growth.

It is generally recommended that for proper conditioning the castings should be held at the specified temperature for approximately 60 min. for each inch of thickness of the largest section of the casting. According to investigation, the factors of holding time and mass are of greater importance than temperature. With larger sized castings it obviously becomes more difficult to refine the structure.

It is quite evident that with castings of large size the only alternative is slow cooling, so that the rate of cooling is hardly a matter of choice in these cases. The rate of cooling may induce either a network or a Widmanstätten structure, depending upon the speed at which the steel is cooled through the critical range. Slow cooling will induce the Widmanstätten and more rapid cooling will promote the network structure. Since the initial rate of cooling during and after solidification has a marked influence on the structure and since different types of structures will be encountered, it becomes necessary to modify the treatment accordingly. Another factor that must be recognized is that the larger the casting size, the slower will be the solidification; hence, the greater the degree of dendritic segregation. The greater the segregation, the longer will be the heat-treatment time necessary for diffusion and, unless the steel is resistant to grain growth, the coarser will be the grain structure.

Since it is difficult to bring about the optimum conditions of rapid and complete diffusion and fine grain size in a single operation, a double heat-treatment is often employed. The double treatment consists in first heating to a relatively high temperature, 1600 to 2000°F., followed by normalizing and a second heating at a temperature slightly above the critical range. These treatments may then be followed by a tempering or spheroidizing treatment.

The first temperature used will effect homogenization and will vary with respect to the degree of original segregation. The segregation will, in general, increase with increase of section, so that the temperature employed will depend upon the mass. The structure, following this treatment, will still be quite coarse, so the second treatment is effected at a temperature just above the tritical, to refine the grain. If the homogenization in the first creatment has been properly carried out, the second treatment becomes a problem of forming as many crystals as is possible and preventing their further growth after forming.

Oil and Water Quenching.—Liquid quenching of steel castings involves no unusual procedures. If exceptional physical properties are specified for cast steels of a given analysis, the procedure would be to heat above the upper critical range, quench in water or oil, and temper to the required physical specifications. The recommended quenching and tempering practice is found in Table 2-XIV.

Where minimum distortion or maximum properties are desired, it is recommended that the castings be fully annealed or normalized before the quench, although this practice is not always essential. This treatment results in a finer and more uniform structure.

TABLE 2-XIV.—QUENCHING AND TEMPERING OF CARBON-STEEL CASTINGS*

Carbon range	Quenching			Tempering	
	Temperature of casting, °F.†	Time at heat	Medium and its temperature, °F.	Temperature, °F.	Time at heat
0.15–0.30	1600–1650	¼ hr. for every inch of diameter or thickness	Water at 70	800–1250	At least 2 hr.
0.30–0.40	1550–1600	As above	Water at 125	800–1250	At least 2 hr.
0.40–0.60	1500–1550	As above	Water at 125 or oil	800–1250	At least 2 hr.

* A.S.M. "Metals Handbook," p. 961, 1939.

† These temperatures are based on the assumption that the castings were first annealed.

TABLE 3-XIV.—PROPERTIES OF MEDIUM CARBON STEEL CASTINGS*
(Average of 80 tests)

Properties	As cast	Annealed	Water-quenched and drawn	Per cent change from unannealed as unity	
				After full annealing	After quenched and drawn
Ultimate strength, p.s.i.	74,100	75,000	80,000 to 125,000	+1.1	+12.6
Yield point, p.s.i.	37,100	41,500	50,000 to 97,000	+12.0	+53.8
Elongation in 2 in., per cent	19.5	24.5	30 to 10	+30.8	+61.5
Reduction of area, per cent	31.0	46.2	65 to 20	+51.7	+124.1
Brinell	160	145	155 to 250	−8.3	+2.5
Izod	16	20	48 to 15	+23.5	+158.8
Endurance ratio	0.40	0.44	0.50	+10.0	+25.0

* MITCHELL, H. A., Getting the Most Out of Steel Castings, *Iron Age*, **126**, 914 (1930).

Where the castings are of such size and of such design that cracking would result if they were kept in the quenching bath until cold, it is important that they be removed from the quenching medium while still warm. The temperature of their removal should be regulated to suit the type of casting and the physical properties desired.

Table 3-XIV compares the improvement in physical properties from the unannealed, to the annealed, to the heat-treated state. The heat-treated castings in this case are typical ones used to replace annealed castings and have been quenched in water from 1650°F. and tempered at 1300°F. The percentage changes are based on the unannealed or as-cast results as unity.

Tempering or Drawing.—Most castings in the full-annealed condition do not require a drawing treatment. Some castings of lower carbon content also will give excellent service without a draw after normalizing. In case of higher carbon and alloy steels, and for intricately designed castings, a tempering treatment is essential. This treatment brings about the relief of the stresses set up by the rapid cooling and improves the ductility and impact resistance, with some sacrifice of strength.

After the castings have been quenched, they should immediately be tempered to the temperature that will produce the specified physical properties. The furnace practice should be so regulated that the tempering furnace is at the temperature of the castings when they are taken from the quenching bath. The temperature ranges for tempering, as is shown in Table 2-XIV, will vary between 800 and 1250°F., depending upon the physical properties desired.

Alloy-steel Castings.—The benefits of adding alloys to steel, to be discussed in more detail in a later chapter, are obtained through their effect on response to heat-treatment. Full annealing fails to take advantage of the benefits that might be obtained and hence is seldom used except as a pretreatment followed by normalizing or quenching, or when machining operations are required.

The decision regarding whether an alloy casting should be normalized or quenched is determined by the type of steel and the use to which it is to be applied. Some alloy steels require normalizing, while others require quenching and tempering to attain the optimum combination of physical properties specified.

Suggested Questions for Study and Class Discussion

1. How is the cast structure in steel castings refined? Compare this with the methods of refinement used in the production of rolled and wrought products.

2. Discuss the practical heat-treatments in the accompanying tabulation, with respect to the reason for using such treatment and to the final structures obtained.

Class of castings	Carbon, per cent	Treatment	Time at heat, hr.
Coupling boxes:			
Blooming and plate mills	0.23–0.28	Anneal, 1625–1650°F.	10–12
Blooming and billet mills	0.50–0.60	Anneal, 1550–1575°F.	6
Pinions:			
28 in. blooming mill.....	0.45–0.55	Anneal, 1575–1600°F.	30
44 in. blooming mill.....	0.45–0.55	Anneal, 1575–1600°F.	48
Rolls:			
44 in. blooming mill.....	0.50–0.60	Anneal, 1550–1575°F.	36
40 in. 3-high blooming mill.		As above—top roll not to be annealed	
28 in. 2-high blooming mill..................	0.50–0.60	Anneal, 1550–1575°F.	30
Driving wheel center on locomotive.	0.35	1550°F. quench in water	Up 4 Hold 4
		1000°F. draw	Up 4 Hold 4

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CHAPTER XV

CASEHARDENING OF STEELS

In the design of machinery, the selection of a ferrous metal with physical properties to provide the strength demanded and at the same time provide a surface hardness that will withstand wear and abrasion is of considerable importance. As has been observed in previous studies, the cheaper ferrous alloys do not possess both of these characteristics unless they have been subjected to some form of heat-treatment. Fortunately, iron, when held at an elevated temperature, has the property of readily absorbing such elements as carbon and nitrogen, which have the property of imparting hardness to the metal. It is possible, therefore, to select a ferrous material of sufficient internal strength and subsequently to introduce these hardening elements into its surface to form a so-called *case*, and thereby produce a finished product with the specified properties of internal strength and surface hardness.

Casehardening is, therefore, applicable and most suitable for parts that are subjected to shock and abrasion or in places where local hardness is required. It is also applicable for parts so complicated that they are likely to warp and/or crack when made from high carbon steels. Casehardening is applied to tools, gears, bearings, automobile parts, etc.

Straight high carbon steels, although they do resist wear and have a high surface hardness when they have been heat-treated properly, do not possess the desired toughness or ductility and are expensive to machine. For moving parts of machinery it is essential that a stock be selected that combines "core" toughness and ductility with the ability to absorb carbon and/or nitrogen to yield the hardness of surface of the high carbon stocks.

The inexpensive low carbon steels serve admirably, in that they have the necessary softness and toughness; and by casehardening the desired surface hardness may be imparted. With proper heat-treatment, the finished product will have a surface that is resistant to wear, while the original soft center will withstand impact and fatigue.

The casehardening processes—carburizing, nitriding, and cyaniding—will now be studied in detail.

CARBURIZING

The casehardening process known as *carburizing* is a treatment for ferrous materials, which in its simplest form comprises two essential stages: (1) impregnation of the surface of the material with sufficient carbon to produce a surface that, after subsequent heat-treatment, will have higher hardness and higher wear resistance than the steel would be capable of without the treatment; and (2) heat-treatment to develop the desired properties in the carburized surface areas (case) and the core. The "case" is that portion of a carburized iron-base alloy article in which the carbon content has been substantially increased. The "core" is that portion of a carburized iron-base alloy article in which the carbon content has not been substantially increased.

In order that a case rich in carbon may be produced by carburizing, we must consider four main factors in the operations: (1) the solvent or the steel, (2) the compound capable of delivering the carbon, (3) the temperature, and (4) the time of contact between the carburizing agent and the steel.

Carburizing Steels.—The principal requirements of carburizing steels are that (1) they should be able to absorb carbon at the required carburizing temperature at a reasonable rate and produce a uniform case of sufficiently high carbon content; (2) they should possess sufficient strength and toughness after carburizing and heat-treating to meet the service requirements; (3) they should heat-treat in a quenching medium demanded by the section with a minimum amount of distortion; and (4) they should be easily machinable.

In the plain carbon steels that are satisfactory for a wide range of applications, the maximum carbon content is generally 0.25 per cent. With carbon content below 0.15 per cent, the steels will have high ductility and good shock-resisting properties but will have low core strength and poor machining qualities. Steels of from 0.15 to 0.25 per cent carbon will, on the other hand, be somewhat lower in ductility and shock-resisting properties but will have higher strength and better machining properties. In these steels the manganese will range from 0.30 to 1.00 per cent, with phosphorus 0.045 per cent maximum and sulfur 0.055 per cent maximum.

Where the service duties of the steel are critical, certain alloying elements such as chromium, nickel, vanadium, and molybdenum are used, either alone or in combination, to give increased strength and wearing qualities, which the straight carbon steels are unable to produce. The carbon content will, in general, be as specified for the straight carbon steels.

Grain size in both the carbon and the alloy steels is of utmost importance since this factor affects the machinability, the suitability of heat-treatments, the distortion during hardening, and the physical properties. The coarse-grained steels are found to machine more easily, harden deeper, and—because of this—distort more during quenching; and they are heat-treated most satisfactorily to produce maximum toughness by the double treatment.

The fine-grained steels do not machine so readily as the coarse-grained steels. However, this can be corrected by normalizing above the grain-coarsening temperature. They do not harden as readily as coarse-grained steels and may show a tendency to produce soft spots, particularly with the plain carbon steels; they are shallow hardening and, as a result, distort less; and, since they do not coarsen at the carburizing temperature, they can be heat-treated to produce a tough core by a single treatment or direct quench.

The Carburizing Agent.—The carburizing materials are divided into three classes: solids, liquids, and gases. By the very nature of the process it is essential that they shall have a ready supply of carbon. In the solid type the parts are placed in suitable containers, surrounded by the carburizing material, and then sealed so that the gases generated, upon heating to the carburizing temperature, will not escape. The carbon from the gases is then absorbed by the material. In gaseous carburizing, the parts are placed in a suitable container, sealed, and held at a given temperature while the gas of definite composition is circulated through it. The liquid carburizers are cyanogen compounds, the parts being immersed in the molten liquid. This last process will be discussed in more detail under the section on cyaniding.

Solid or Pack Carburizing.—Carburizers of the solid type are, perhaps, more generally used than are those of the other two types. Many satisfactory compounds are available, of which the charcoal-coke type is probably the most widely used. This type consists of a mixture of hardwood charcoal and petroleum coke, energized with barium carbonate, calcium carbonate, and sodium carbonate. In general, the usual carbonaceous ingredients used consist of charcoal, charred bone, charred leather, bituminous coal, and different kinds of coke. The energizers or catalysts for the carburizing process, which are quite important, as will be seen later, consist of the alkaline earth and alkaline metal compounds, which have been found to be more effective, in general, than the compounds of other groups. For further reference to the compositions of various solid carburizers the readers are referred to the A.S.M. "Metals Handbook" and the many commercial bulletins published by companies handling this material.

The desirable features of a good carburizer, according to Wyzalek,[1] are (1) the rapidity at which it transmits the carbon to the steel at the temperature used, (2) lasting qualities, so as to ensure uniform carburizing performances, (3) low specific gravity, *i.e.*, minimum weight per cubic foot, so as to give maximum volume per ton purchased, (4) minimum shrinkage when subjected to the carburizing temperature, thereby enabling the containers to be filled to maximum capacity, (5) good thermal conductivity to aid uniform heating of the charge, and (6) freedom from impurities that would contaminate the steel through either absorption or deposition.

Therefore, carburizing compounds should be checked for chemical composition, which will indicate the definite percentages of compounds present and the limitations that should be placed on such undesirables as moisture, sulfur, and inorganic compounds. Physically the carburizer should be of definite particle size and of definite weight per unit volume, and should have low shrinkage loss per run.

Packing and Containers.—Packing pots or containers are of more importance to satisfactory carburizing than that of merely being containers to hold the carburizer and parts. These pots are

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[1] Wyzalek, J. F., Carburizing Practice, *Metal Progress*, **27** (No. 2), 19–23 (1935).

made of cast iron, cast and pressed steel, and special heat-resisting alloys. Long life is, of course, a factor in the type used, but the scaling and warping characteristics are of equal importance. The heat-resisting alloy containers are used quite generally because they are nonscaling in comparison with those made of iron or steel, which will scale after a few runs and thereby affect the conduction of heat to the interior. This would affect the temperature uniformity of the contents, which will in turn affect the uniformity and depth of the case. General design of the container is another important requirement. The design should be such that there will be uniform conduction of the heat to the charge. The size of the pot, whether it be round or square, should be governed by the size and shape of the parts. Small parts should not be placed in an extremely large box, as it must be remembered that with a large box there will be a considerable temperature gradient between the furnace temperature and the center of the pot. This means that a small piece in the center of the pot will lag behind the temperature attained by parts closer to the sides and will, therefore, have a thickness of case different from those closer to the sides.

Another important factor in uniform carburization is the packing of the products. For uniform heating and satisfactory carburization, we must take into account the necessary spacing of the parts, which really is a problem of the amount of carburizer to be used. This will vary with respect to the shape and size of the pieces handled and with the operating conditions and metallurgical requirements of the finished parts. In general, the pieces should be placed 1 to 1½ in. from the sides of the box and should be so separated from each other by the compound that the carburizing gases can circulate freely. In no case should the parts touch the sides of the box, nor should they touch one another. After final packing, the parts should have been covered on all sides by a suitable amount of carburizer. The lid is then placed on the box and the edges are carefully sealed with fire clay or asbestos cement. It is obvious that the furnace gases must not penetrate into the box, lest the carburizer burn and cause local overheating, which will ruin the work. The boxes do not necessarily have to be altogether airtight, but they should be reasonably so. Here, again, is where the problem of distorted boxes becomes a factor.

Selective Carburizing.—It is often desired to carburize some parts of a piece without carburizing others. Selective carburization is applied in special cases where it is desired to retain soft surface portions after hardening, to permit machining, polishing, or straightening, or to increase the toughness of the piece. For example, for such work as automobile camshafts it is desirable to carburize only the bearing and wiper surfaces and to protect the shaft for subsequent treatment, such as is mentioned above.

The most widely used method is the application of a good copper plate by electrolytic means, sometimes to the entire piece, after which the plate is removed by grinding from the portion to be carburized. In other cases, where it is difficult to remove the copper plate, a "stop-off" material is used to prevent the deposition of copper on the parts to be hardened. Stop-off materials are composed of beeswax and rosin or asphaltum, dissolved in turpentine. Another method is, when machining, to leave excess stock on the portion desired soft. After carburizing, the piece is box cooled and the excess stock is machined off below the case, leaving a soft surface after the heat-treating operations. Other methods are merely the application of some material to protect localities against carburizing.

Furnaces.—Recalling again that the two general types of furnace are the batch and continuous types, let us see what type has been found most successful in production carburizing operations. On large-scale production of medium and large-sized parts, the continuous type of furnace has proved the best, both for low operating cost per part and for uniformity of case, this latter probably being due to the fact that all containers are subjected to identical treatment. The rotary-hearth or rotary-batch type would be most desirable for small parts. The batch type, of which the car-bottom furnace is an example, is most suitable for reasonably small quantities of medium and large-sized parts.

Heating.—The temperature of the pots may be judged in two ways: (1) by estimation with the eye, and (2) by placing a thermocouple inside the pot and one in the furnace so as to make the schedule of heating more constant and easier for the operator. It is obvious that, where the pots are so small as to make it impractical to use a thermocouple in the box, the operation should be controlled by a thermocouple in the furnace. In heating the pots, regardless of what means of control is used, it is essential that the heating shall be relatively slow, so that all the pots will reach the carburizing temperature as nearly as possible at the same time. It is essential, then, that a schedule of heating shall be standardized for each total furnace load with respect to the size of the pot, and that this schedule shall be strictly adhered to by the operator.

Wyzalek states[1] that an over-all variation of 25°F. in the carburizing zone is permissible and will give satisfactory carburization. After a definite temperature is established, a deviation of 10° plus or minus is recognized as good practice.

The temperature of carburizing will depend largely upon the requirements of the finished part. In general, temperatures of between 1600 to 1750°F. will give a satisfactory rate of penetration. Where a shallower case is desired and distortion becomes a serious problem, particularly on small flat parts, temperatures of 1375 to 1450°F. are more suitable.

Theory of the Carburizing Process.—When the low carbon steels or other ferrous materials are heated to the carburizing temperature, there is complete transformation to austenite. At the carburizing temperature gamma iron is capable of holding in solid solution an amount of carbon approaching the saturation limit of 1.7 per cent. When a source of carbon, such as carbon monoxide, is brought into contact with the steel in this condition, there will be a transfer of carbon from the gas to the steel. There is a difference of opinion as to the exact form of the carbon: some feel that there is an interaction of the carbon monoxide and the iron to form iron carbides, which are then absorbed by the steel; others feel that the carbon being brought to the surface of the metal as carbon monoxide will, in the presence of austenite, break down to form atomic carbon, which will diffuse as such in the iron. For our present purpose of discussion let us assume that the action involving the formation, absorption, and diffusion of iron carbides is the essential part of the carburizing process.

The carburizing compound is in reality the gas producer. In other words, the carbon monoxide is the result of the reaction between the carbonaceous compound and the oxygen from the air, the air in this case being the air occluded with the compound

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[1] Wyzalek, J. F., Carburizing Practice, *Metal Progress*, **27** (No. 2), 19–23 (1935).

particles. The CO thus formed will give up its carbon to the iron according to the following equations:

$$2CO + 3Fe \rightleftarrows CO_2 + Fe_3C \text{ (or Fe } - \text{ C solid solution)} \qquad (1)$$

Without a further supply of CO this reaction between CO, CO_2, and austenite would soon reach equilibrium and carburizing would stop. Herty states[1] the CO: CO_2 ratio must be above 24 for this reaction (1) to proceed and also that the rate of penetration must be about 0.01 to 0.02 mm. per hr.

Since, at the carburizing temperature, there is always a large amount of incandescent carbon present, the CO_2 will be constantly reduced, so that there will be a continuous supply of CO according to the reaction:

$$CO_2 + C \rightleftarrows 2CO \qquad (2)$$

This cycle continues to repeat itself, and the process of carburizing will continue as long as the proper temperatures are maintained and the CO: CO_2 is kept high enough. In this case the carbon will be absorbed, the concentration of carbon will be built up at the surface and will then move inward according to the simple laws of diffusion of heat or dissolved substances, *i.e.*, from a region of high concentration to one of lower concentration. The carburizing phase of casehardening is, thus, one of absorption at the surface, plus inward migration.

Since the usual carburizing compounds have incorporated in them one or more of the carbonates of the alkaline or alkaline-earth metals it is generally considered that they may react according to the two following equations:

$$M_xCO_3 + \text{heat} \rightarrow M_xO + CO_2 \qquad (3)$$

The M is any metal of the alkaline or alkaline-earth group. The CO_2 in this case will react with the hot carbonaceous packing, as in reaction (2), and will result in increasing the concentration of available CO.

It is also believed that the carbonate reacts with the hot carbon:

$$M_xCO_3 + C \rightarrow M_xO + 2CO \qquad (4)$$

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[1] **Herty, H., and others, Abnormality in Case Carburized Steels, *Cooperative Bulletin* 45, Carnegie Institute of Technology, and Mining and Metallurgical Advisory Boards, 1929.**

and that the oxide thus formed will act as a catalyst between the reaction of the CO and the iron.

Gas Carburizing.—From this discussion, it would be natural to postulate the question: If CO gas is the actual carrier, why not introduce CO itself in a proper container with the work held at the proper temperature and thus carburize? Investigation has shown that CO will act in this capacity but the action is relatively very slow in comparison to some hydrocarbons, unless it is in the presence of large masses of incandescent carbon. Therefore, for proper gas carburizing, we should use a gas that will supply carbon at a greater rate. For this purpose there is available a number of gases and mixtures, ranging from ordinary carburetted water gas to such compounds as propane or butane. The last named two gases present an added difficulty in that, when heated, they do not break down directly with the liberation of carbon and hydrogen but form a number of intermediate products, among which are some resinous substances that will form coke. The formation of coke on the surface of the work to be carburized interferes with penetration and hardening and, on further treatment, will result in a shallow case and insufficient hardening. The problem in gas carburizing then becomes one of control of the rate of carbon deposit. Several methods have been suggested and put into practice for such control. The control is essentially one in which the active richer hydrocarbon gases are diluted by relatively inert (to the carburizing operations) gases, such as carbon monoxide, lean city gas, or nitrogen, so that the carbon is deposited on the work at a much reduced rate and within limits that are controllable. The proportion of these gases can be varied over a considerable range without any noticeable change in the production of a satisfactory case, so that commercial control is comparatively easy.

Turbulence and velocity of the gases in contact with the work furnish a very important factor in obtaining case uniformity in gas carburizing. The velocity of the gas must be great enough so that there is circulation of the gas throughout the retort and thus cause every part to be swept by the gas.

Furnaces.—The batch type of retort furnace and the continuous type of furnace both may be used successfully with controlled-gas atmosphere. In the production of low carbon cases, the batch-type retort furnace functions quite successfully, owing to the fact that there can be established a cycle in which the carbon is being supplied to the surface for a given length of time at a carburizing temperature, after which the gas is cut off and this carbon is allowed to diffuse into the work. The work is maintained at the carburizing temperature throughout the diffusion period. During this time there will be a decrease in the percentage of carbon at the surface, with an increase in depth of the case. With the continuous type of furnace, this type of control for a low carbon case is quite difficult. For the higher carbon case requirements, the continuous furnace is quite satisfactory.

Depth and Carbon Concentration of the Case.—The depth of the case for a given application will be dependent upon the specific service requirements demanded. For example, with small parts, such as fine teeth on small gears, the case must necessarily be shallow so that sufficient core will remain to provide the toughness. For large parts that must maintain heavy, concentrated loads, the case must be deeper, in order to have the strength necessary to resist the crushing forces.

The decision as to whether a light or a heavy case is to be used may be made by considering the following principal situations, which warrant the use of a heavy case:[1]

1. To withstand a heavy load so that the case will not collapse.
2. To withstand a great amount of wear. This applies only if the part can lose dimension without losing its usefulness.
3. Because of grinding to be done after heat-treatment. Excessive finish grinding should be avoided, not only because it comprises wasted heat-treatment, but because the portion removed is probably the most vital part of the case as far as hardness and wear resistance are concerned.
4. To build up the physical properties of the surface.

With modern carburizing procedures (solid compounds), the carbon concentration at the surface of the case will range from approximately 1.05 to 1.20 per cent. The remainder of the case will taper off gradually to the core carbon content. In general, when solid carburizers are being used, and when satisfactory properties are being developed, the usual practice is not to attempt to control the carbon content. This is probably satis-

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[1] Williams, G. T., "What Steel Shall I Use?," A.S.M., Cleveland, Ohio, 1941.

factory for general work since the carbon at the surface has attained the above figure within the first few hours and any additional time merely increases case depth and not maximum surface carbon.

The relation of case depth and carbon concentration of the case is found to be controlled by the following factors:[1]

1. The amount of carbon available to the steel surface; if this is above the limit determined by the iron-carbon diagram (SE line for the given carburizing temperature), this factor is at its maximum.

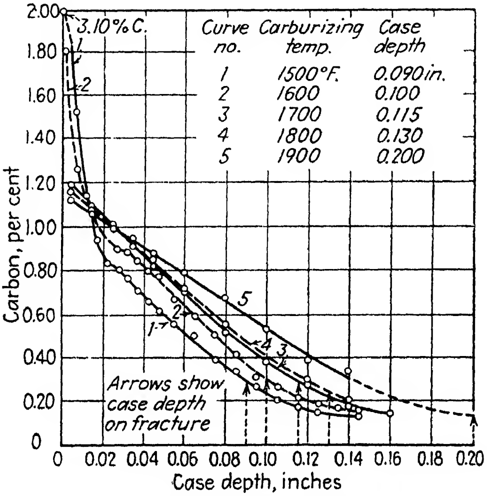

FIG. 1-XV.—Effect of temperature on the carbon gradient in an S.A.E. 3115 steel carburized at 1700°F. in a hardwood charcoal, coke, sodium carbonate compound.

2. Continuity of supply of suitable carbon to the steel surface.

3. Temperature: Carburizing action depends upon combination and inward migration, and these are facilitated by temperature increases but not to the same degree. The effect of carburizing temperature on the carbon gradient when carburizing an S.A.E. 3115 steel in a hardwood charcoal-coke mixture energized with sodium carbonate is shown in Fig. 1-XV.[2] It is seen from

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[1] WILLIAMS, G. T., Case Hardening, *Industrial Heating*, **6**, 880–894, 956–957 (1939).

[2] SCHLUMPF, R. W., The Mechanism of Carburization, "A.S.M. Metals Handbook," p. 1039, Cleveland, Ohio, 1939.

the curves that at the low carburizing temperatures there has been a high build-up of carbon at the surface, which is far in excess of the equilibrium value for saturated austenite. This is due to the fact that at the lower temperatures, the ability of the carbon and iron to combine is less than the rate of migration (diffusion), so that there is a rapid build-up of the carbon at the surface. At the higher temperatures, the balance between carbon build-up and diffusion results in a surface carbon concentration closely approximating the equilibrium value for saturated austenite, as is indicated by the *SE* line of the iron-carbon diagram.

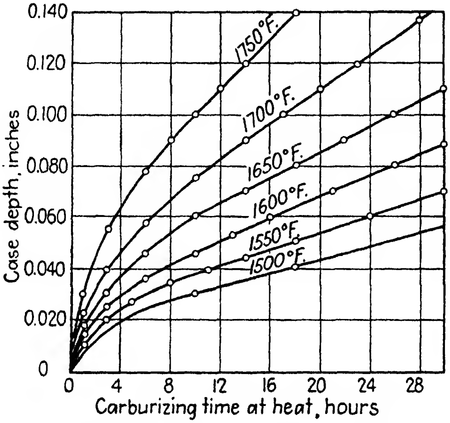

Fig. 2-XV.—Relation of time and temperature to carbon penetration. Carburized in vertical gas retort using natural gas (95 to 98 per cent methane). Measurements made on a triangular test specimen. Steel, S.A.E. 3115. (*Courtesy of Bethlehem Steel Company.*)

Most carburizing with gaseous and solid agents is performed at temperatures of from 1650 to 1750°F., while liquid media are usually used at 1500 to 1700°F.

The relation of time to carbon penetration or case depth is illustrated in the set of curves (Fig. 2-XV)[1] that have been determined from tests run in a production vertical, stationary type of gas retort using natural gas at atmospheric pressure. These curves approximate closely the rates obtainable, regardless of the medium used. The curves show that there is a rapid initial rate of carburization and a later decreasing rate, as is indicated

[1] Schumpf, R. W., The Mechanism of Carburization, A.S.M. "Metals Handbook," p. 1041, Cleveland, Ohio, 1939.

by the varying slope of the curve and the extreme importance of temperature.

As can be seen from the time-temperature-penetration curves, there is a time beyond which the rate of penetration reaches a low and constant value. As a result, the case-depth specifications are usually kept low enough so that the time is not excessive, the most common case depth ranging from 0.045 to 0.060 in. Any increase in case depth beyond this normal figure is expensive to attain and should be specified only when demanded by service requirements. Any decrease in depth may be permitted when the service is light or when the core properties are improved by the use of higher carbon or alloy content.

4. Steel composition: Since migration is a vital factor, any variations in steel structure caused by various elements, added deliberately or accidentally, may affect the rate of migration.

Microstructure.—The microstructure of a carburized, unhardened S.A.E. 1015 carbon steel is shown in Fig. 3-XV[1] *A* and *B*. *B* of the figure shows the hypereutectoid zone that is adjacent to the edge of the specimen; the network of carbide around the grains is apparent. The dark area at the extreme top of *B* is the eutectoid zone in which there is neither excess cementite nor ferrite. *A* of figure 3-XV represents the core zone, which has been unaffected by the carburization and is composed of ferrite and pearlite. A composite photomicrograph of an entire gradient zone is shown in Figs. 5-XII to 12-XII, inclusive. The hypereutectoid zone plus the eutectoid zone plus a few thousandths of an inch of the hypoeutectoid zone constitute the hardenable case and the depth that will be found on fracture.

There is some difference of opinion as to what constitutes an ideal case, but it is generally agreed that the case should be gradual, much the same as has been illustrated. It is sometimes found that there is a sharp line of demarcation formed between the different zones of the case, *i.e.*, there may be a concentration or an agglomeration of the cementite in large patches near the surface of the part and also an agglomeration of the ferrite deeper in the case where the carbon is under the eutectoid

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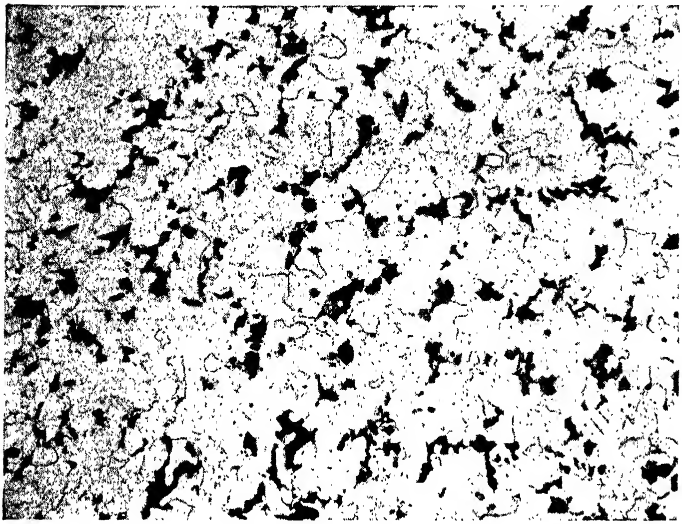
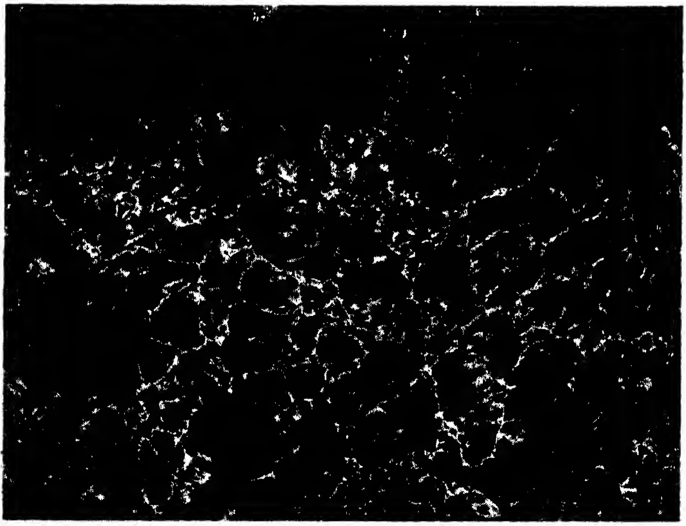
[1] The microstructure shown was produced by pack carburizing, but it is similar to that obtained by gas carburizing or carburizing in a salt bath. With the salt bath, however, the penetration will, in general, be much less and the maximum carbon content attained in the case will be lower.



A



B

FIG. 3-XV.—*A*, carburized 1015 carbon steel. Core × 100. Nital etched. *B*, carburized 1015 carbon steel. Case × 100. Nital etched.

composition. If this condition, known as *exfoliation*, should persist in the finished part, it will be likely to cause chipping or spalling at this demarcation line when in service. Such a concentration is due in part to high carburizing temperatures, which tend to build up and concentrate the carbon at the outer surface of the part with little gradation of the carbon from that zone toward the core, and to the effect of slow cooling from the carburizing temperature. This effect is corrected by changing the carburizing cycle or it may be alleviated by subsequent heat-treatment. In gas carburizing—particularly in the use of the retort furnace—the diffusion period ensures proper gradation of the carbon from case to core. In pack carburizing, a similar effect may be obtained by carburizing at a high temperature, 1700 to 1750°F., for a given time, thus building up the carbon concentration at the surface, and then dropping the temperature to 1600 to 1650°F., slowing up the absorption of the carbon, and allowing the carbon already present to diffuse and produce an even gradation.

Heat-treatment.—Before discussing heat-treating, let us visualize the structure present at carburizing temperatures. The entire part at that temperature will be in the austenitic condition. There will be the case portion, with high carbon content, the carbon being in solution as Fe_3C (or carbon in solution), and the core of low carbon content. On cooling, the different zones will precipitate their constituents as their respective critical ranges are reached. At room temperature the carburized layer will consist of pearlite and cementite—the proeutectoid cementite either being in the grain boundaries outlining the original austenite grains or having become agglomerated in large patches of pearlite, and pearlite and ferrite.

This description of the constituents present in a carburized steel should make it quite clear that the heat-treatment of a carburized steel will not be the same as that of a homogeneous steel. In this case we are dealing not with one steel but with a group of steels varying from, say, 1.20 carbon down to 0.15 carbon. Any heat-treatment imposed on this multiplex steel for refinement of its structure must be made with due regard to its effect on the entire carbon range.

In general, the following suggested heat-treatment will apply for the carbon steels as well as for the alloy steels, with the exception that the actual temperatures may be changed somewhat, because of the influence of some of the alloys on the position of the critical range.

In general plant practice there are two factors that must be stressed very strongly. The first is the composition of the steel to apply in the case of a given part; *i.e.*, if strength or wear is demanded, or both, the composition will be varied so that subsequent treatment will give the physical properties required. The second is the austenitic grain size of the steel applied.

The subsequent treatments will be discussed with reference to both factors.

Coarse-grained steels (2- *to* 4-*g. size*).

1. Carburize at 1600 to 1700°F.
   Cool in box.
   Reheat to above the critical of the case Ac_1.
   Quench in oil or water.
   Temper 250 to 500°F.

This treatment is used in instances where wear is of utmost importance and case strength of little or no importance. This treatment will, therefore, give a hard martensitic case, the precipitated free cementite being unaffected. The core structure will also be unaffected. The original coarse-grained structure, at the low temperature, will be refined in the case but will remain the same in the core. In this condition the part is spoken of as having an unrefined core with a refined case. In all subsequent treatments the casehardened parts are, as a final treatment, tempered at low temperatures 400°F. and below for relief of stresses, and at higher temperatures above 400°F. for increase in toughness and a decrease in surface hardness.

2. Carburize at 1600 to 1700°F.
   Cool in box.
   Heat to above the upper critical of the core.
   Quench in oil.
   Heat to above the lower critical of the case.
   Quench in oil.
   Temper 250 to 500°F.

This treatment will produce maximum resistance to wear in combination with increased core strength. The first reheat and quench produces maximum refinement and best physical properties of the core, while the second reheat and quench results in refinement of the case.

An alternative to the above treatment—and one that is probably preferable—is the elimination of the first reheat by quenching directly from the box into oil and then following through as before. With correct furnace practice, this method assures uniform heat in the parts and eliminates any further variables that might be encountered by the reheating. The same properties are obtained by this method.

Fine-grained steels (5- *to* 7-*g. size*).

Since steels of the fine-grained type can be purchased now, it is possible to quench directly from the box and still have a fine-grained case and core that will yield maximum strength and wear. This method of direct quenching is made possible because of the fact that these controlled steels do not coarsen at normal carburizing temperatures.

Carburize at 1600 to 1700°F.

Quench from that temperature into oil or water.

Temper 250 to 500°F.

Which one of these treatments suggested is the best for a given part depends not so much on the quench to refine the core and case as it does upon the microstructure of the case, its depth, and the characteristics of the transition zone that will give longest life in service. Any treatment that will give such a structure will be satisfactory.

In all the treatments discussed, the general precautions regarding heating, cooling, furnace atmosphere, quenching media, design, etc., must be carefully observed.

H. W. McQuaid[1] lists some of the fundamentals of the effect of treatment on the over-all properties of casehardened steels. Those of his list that are pertinent to this discussion are the following:

1. The deeper the case, the greater is the tendency to distort.
2. The more nearly the carbon content of the case approaches eutectoid, the higher is the over-all strength of the carburized piece.

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[1] McQuaid, H. W., Selecting the Proper Hardening Treatment, *Metal Progress*, Vol. **27** (No. 3), 25–30 (1935).

3. The higher the carbon in the case, the greater is the hardness. It should be mentioned here that the surface hardness, as indicated by the scleroscope, the Rockwell, or the Brinell test, depends on the microstructure and analysis being tested and is not always related to the file hardness and service life of the part.

4. For maximum resistance to abrasion, together with maximum stability of case, the higher carbon, fine-grained types containing an alloy are the most satisfactory, particularly when directly quenched.

5. For maximum case strength, the lower carbon, fine-grained types are recommended.

For a detailed discussion of physical properties developed in both the case and the core of hardened steels, the reader should refer to the reference in the footnote.[1]

CYANIDING[2]

Cyanide hardening may have three meanings: (1) the heat-treatment of a straight carbon or alloy steel through the use of a fused salt bath containing cyanide to prevent decarburization; (2) the treatment of a low carbon steel in such a bath with the idea of incorporating on the surface of the steel, a file-hard wearing surface or case, a practice customary in the regular "cyaniding" operation used on commercial steels of that classification; and (3) as a cyanide "reheat" used for treating oil-hardening alloy steels containing about 0.35 per cent of carbon to obtain a shallow, file-hard, wear-resisting surface layer. In the last instance, the bath used for this purpose will generally contain from 25 to 35 per cent of cyanide. The first and third examples are devices of recent origin, but the second has been in practice for years and with only minor improvement.

Cyanide Compounds.—The earliest cyanide hardening, practiced by the "old-school" blacksmiths, made use of the yellow prussiate of potash (potassium ferro cyanide). The steel (almost any grade) was heated to a "cherry-red" heat and then sprinkled with the powdered "prussiate" or else rolled or dipped in the powder. It was then reheated, resprinkled, reheated, and

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[1] McMullan, O. W., The Physical Properties of Case Hardened Steels, *Trans., A.S.M.*, **23**, 319–381 (1935).

[2] A.S.M. "Metals Handbook," pp. 1063–1067, Cleveland, Ohio, 1939.

quenched, the number of applications depending on the depth of case desired.

Next, we find mention of the use of a cast-iron pot containing potassium cyanide (KCN). This was heated until the salt was fused. The temperature was then raised to a dark cherry-red heat and the steel was dipped or immersed in the fused salt and held there until the operator throught that the piece would become file-hard upon its removal and quenching in water. Owing to the expense of the potassium cyanide, the process was limited mainly to toolroom or machine shop, nonproduction hardening.

With the advent of the First World War, potassium cyanide sold at a premium. Luckily, at just about that time, the atmospheric fixation of nitrogen afforded a very much cheaper and at the same time a much more effective substitute for the potassium cyanide—*i.e.*, the sodium cyanide in wide use today. This is an active chemical compound containing sodium, carbon, and nitrogen, with the chemical formula NaCN. In its pure commercial state, it is a white crystalline product and has a melting point of 1040°F.

Since cost is a determining factor in any heat-treating process, pure sodium cyanide is seldom used. It has been found that a bath of fused pure sodium cyanide loses its strength rapidly. Research has also determined that a fused salt bath containing 30 per cent of sodium cyanide, 30 per cent of salt, and 40 per cent of sodium carbonate or soda ash is just as effective in producing a case as is the pure sodium cyanide and costs about one-third as much. Thus it is fairly accurate to say that the majority of cyanide hardening today is carried out through the use of a 30 per cent sodium cyanide.

Plant Practice.—Usually, a plant doing any considerable amount of cyanide hardening will purchase two grades of sodium cyanide, the 96 to 98 per cent grade (pure commercial sodium cyanide) and the 30 per cent grade. Plant practices are usually standardized so that the fused cyanide baths are maintained at a definite cyanide concentration. Therefore, when a definite mass of steel is immersed for a definite length of time in a fused sodium cyanide bath of constant and definite composition at a definite temperature, definite case depths and performance of the steel so treated can be predicted accurately. This is just an application of good common sense coupled with accurate metallurgical control.

Cyaniding Procedure.—In operation, the pot is filled with the 30 per cent sodium cyanide, melted down, and a sample is taken from the bath and analyzed. It is customary to maintain the cyanide concentration or strength of the bath as close as possible to 25 per cent NaCN. The steel being treated fixes the temperature at which the bath should be maintained. With this temperature established, the rate of volatilization is determined by frequent analysis. Once the rate of volatilization has been established, the strength of the bath is maintained constant through frequent and regular additions of 96 to 98 per cent strength sodium cyanide. "Dragout," or loss of bath volume due to pulling the heat from the pot, is replenished to keep the pot at constant level by the addition of the regular 30 per cent sodium cyanide mixture. These practices should be established for each given set of operating conditions.

The Cyanide Reheat.—The foregoing discussion deals with the cyanide bath designed for the case carburization of steel. Where the bath is to be used as a heating medium only, for protecting the steel against decarburization, the bath should preferably analyze not over 10 per cent NaCN. Case-carburized steel, oil-hardening gears, and tool steels may be heated in such a bath for hardening. The pieces are held in the bath only long enough to ensure their being raised to the proper quenching temperature. During this interval they become thoroughly and uniformly heated, without decarburization or scaling, and they will, in addition, gain just a little in surface hardness, owing to a very slight carburizing or nitriding action of the bath.

The cyanide bath offers an ideal heat-treating medium for finished machined work. This treatment greatly improves the surface hardness and wear resistance. It will also maintain the finish, and attractive surface colors may be developed when the work is properly quenched.

The cyanide reheat is the most widely used treatment for automotive gears. Three processes are in use at the present time. They are designated as the *cyanide dip*, the *wash*, and the *reheat*. In the first two treatments, the work is preheated in a rotary or "hump" furnace and then treated in a cyanide bath for 15 sec. to 4 min. before being quenched. In the cyanide reheat, the cold work is placed in the cyanide bath and treated for 10 to 40 min. These treatments produce a mixed nitrogen-carbon case that greatly increases the wear resistance of the gear. Dynamometer and road tests have shown that cyanide-reheated gears wear more than ten times as long as those that are hardened by the regular oil-treating methods. It has also been demonstrated that in some cases a 15-sec. cyanide dip will eliminate "tooth pitting." Many other articles requiring high surface hardness and wear resistance, such as engraving dies, high-grade end wrenches, high-grade nuts and bolts, etc., are profitably and successfully treated by this method. In general, the shortest time and the lowest temperature that can be used will give the best product at the lowest cost.

The Time Cycle.—The steels most generally adapted to case carburizing are the low carbon, straight carbon steels, preferably under 0.20 per cent carbon, the low carbon nickel and nickel-chromium steels, corresponding to S.A.E. 2315 and S.A.E. 3115. These steels all carburize or cyanide at about the same rate, giving about the same case depth and the same carbon concentration on the surface; but the alloy steels—particularly, the nickel steel—give a very much smoother gradation from case to core. Cyaniding cycles seldom exceed 1 hr. in duration. If case depths are desired deeper than those that can be obtained in a 1-hr. cycle, some other means of case carburization is usually employed.

The cyaniding of steel seldom yields a carbon concentration in the case of more than 0.65 per cent C, yet these steels can be made file-hard upon drastic quenching. Upon first thought, this phenomenon does not seem to be regular. However, if we recollect that one of the decomposition products of a sodium cyanide bath is nitrogen, we will understand that the steel has been subjected to a nitriding action as well as a carburizing action, thus accounting for the extreme file-hardness of the properly cyanided and quenched steel. At 1500 to 1550°F., the surface 0.004-in. layer will have an average nitrogen content of 0.5 per cent; 20- to 40-min. reheats give about the same amount of nitrogen in the surface 0.002-in. layers.

Cyaniding Temperatures.—The cyaniding temperature is usually just slightly above the critical point of the core or original steel. For the steels mentioned above, then, the temperature will range from 1550 to 1600°F., frequently raised to 1625°F., but seldom reaching as high as 1650°F., the usual temperature employed in a solid case-carburizing operation.

Quenching Operation.—Cyanided steels should be quenched directly from the pot. They may be quenched in either water or oil, the selection of the quenching medium being a function of the steel to be treated and of the use to which that steel is to be subjected in practice. If water is used as a quenching medium, it should be as cold as possible. If water agitation is practiced air should not be used as an agitator. If oil is to be used as the quenching medium, it should be a straight-run mineral oil of about 340° flash and 400°F. fire point. The so-called *soluble* quenching oils should never be used, since they contain fatty acids that saponify to a hard soap upon contact with the alkali from the cyanide bath, thus slowing up the quenching and making a general mess of the quenched work, which is hard to clean.

Cyanide Containers. *Cast-iron Containers.*—Cast iron fails prematurely and rapidly as a result of growth. The free graphite of the cast iron grows when subjected to cyaniding temperatures for a considerable length of time and causes distortion or cracking.

Pressed-steel Containers.—Pressed-steel containers are probably those most generally used and, under proper furnace conditions, give good results from the point of view of economy. They are highly resistant to the action of molten cyanide, and are homogeneous throughout, while their cost is moderate and commensurate with the increased life of the container.

Cast-steel Containers.—Cast-steel containers also have a wide application and, when of the proper composition and properly cast so as to be free from casting defects, give satisfactory service.

Cast-alloy Containers.—Nickel-chromium alloy containers have been developed that will give an operating life of several hundred furnace hours. Generally, this type of container is used in electrically heated furnaces and, under such conditions, may give from 5,000 to 7,000 hr. of operating life. As a rule, the premature failure of the cast-alloy containers is due to faulty foundry practice. When they are properly cast, their higher cost is offset by their increased performance.

Cyanide Containers and Furnaces.—The life of any container is shortened by the improper application of heat. As a rule, even physically sound containers will fail from this cause before they will fail from the corroding action of the molten cyanides that they contain. Fuel-fired furnaces should, without exception, have their heat applied near the *top* of the pot, with the furnace vented at the *bottom,* so that the salt will melt from the top down and thereby prevent overheating of the salt at the bottom. In addition to this, in oil-fired furnaces the combustion chamber should be located adjacent to, but outside, the heating chamber. Such a change from the underfired principle has been known to increase the life of pressed-steel pots over 200 per cent. Insofar as pot life is concerned, electric heat applied to the outside of the pot excels all other methods of heat application.

It may be safely stated that the life of any container is in direct proportion to the effectiveness with which the pot is sealed to prevent cyanide from entering the heating chamber. Some furnaces and pots are so designed that it is practically impossible for the molten cyanide to creep into the combustion chamber. In the most common types of furnace, which do not have specially designed seals, a most satisfactory seal against molten cyanide is obtained by placing a ring of dry, powdered fire clay under the pot flange just before dropping the pot into place upon the furnace top.

The most striking instances of progress in cyanide hardening in recent years are the use of the controlled bath of low (25 per cent) cyanide concentration, the design of cyanide pot furnaces top fired and bottom vented, and the improvement of alloy containers to withstand increasingly severe operating requirements. It should be noted that cyanide does not attack sound steel. Improvements that have been made in installations using steel containers have resulted mainly from improved firing conditions, better furnace design, and sounder cast and pressed steel pots. Imperfections in the cast-steel pots, resulting from inclusions, poor casting technique, and improper heat-treatment have, in the past, been responsible for many premature failures. The cyanide bath will dissolve the inclusions and the oxides with which they come in contact but it does not attack the sound metal. Thus, in the case of poor-quality containers, the cyanide has been blamed for attacking the metal, while in reality it was merely disclosing the imperfections in the supposedly sound metal. These developments have proceeded in stages and have tended to stabilize cyanide hardening and make it a more acceptable hardening process.

It should be mentioned that all of the cyanides used in heat-treating are violent poisons. Care should always be exercised against swallowing any pieces of cyanide or breathing in the dust coming from the cyanide cans. Also, cyanide should never be allowed to come into contact with cuts or abrasions of the skin. On the other hand, aside from producing an irritation of the mucous membrane and causing sneezing, the fumes arising from the cyanide pots are not injurious. The fume arises mainly from the oxidation of the sodium cyanide to sodium carbonate in the high-strength cyanide baths and from the volatilization of the salt and the soda ash, if the diluted or lower strength baths are used. As such, the fumes are not usually considered harmful but constitute somewhat of a nuisance and require ventilation equipment for their removal.

NITRIDING

The nitrogen casehardening process as a commercial means of producing an extremely hard surface on steel parts is a recent development. The invention of the process as it is used at the present time is credited to Dr. Adolf Fry, of the Krupp Works, Essen, Germany. This process may be defined as a heat-treating process whereby an extremely hard surface is obtained by heating the steel in contact with a source of nascent nitrogen at temperatures ranging from 850 to 1200°F., usually from 1 to 2 days. The temperature ordinarily employed is below the critical range, at least the critical range of plain carbon steels, and it ranges within 950 to 1000°F. Following the processing, the parts are slowly cooled in the furnace to at least 400°F., after which they may be removed and cooled to room temperature.

Nitriding Steels.—The compositions of the most commonly used nitriding steels are shown in Table 1-XV. (N, preceding the number, indicates nitralloy steels.)

McQuaid and Ketcham[1] have stated that practically any steel can be nitrided, and that, in plain carbon steels, there will be a depth of penetration or case of less than 0.001 in. With the addition of such elements as aluminum, chromium, and molybde-

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[1] McQuaid and Ketcham, Some Practical Aspects of the Nitriding Process, *Trans. A.S.S.T.*, **14**, 719–743 (1928).

num, however, a commercially satisfactory case may be produced. It is now considered that the aluminum and chromium are most effective in contributing extreme hardness to the case, owing—particularly in the case of aluminum—to the amount of nitrogen with which they will combine, as well as to the stability of their nitrides. Molybdenum and vanadium increase the toughness and depth of penetration.

In regard to conditions of service where core strength might be desired in order to resist high-pressure service or where ductility in forming or machining is desired, the various compositions may be shifted; or steels of suitable composition may be used to meet these demands.

TABLE 1-XV*

Elements	N125 (type H), per cent	N125N† (type H with nickel), per cent	N135 (type G), per cent	N135† (modified), per cent	N230, per cent
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.S.M. "Metals Handbook," pp. 1071–1081, Cleveland, Ohio, 1939.

† Aircraft specifications.

Since the usual nitriding temperatures are below the critical temperatures of the steel, the treatment will not change the physical properties of the core or induce warpage or distortion of any appreciable amount. Steels to be nitrided are, therefore, quenched and drawn and machined prior to nitriding. This treatment, of course, postulates that the drawing temperature has been at least as high as the subsequent nitriding temperature; otherwise, there would be a change in the physical properties. A sorbitic structure seems to be most satisfactory prior to nitriding. The nitriding steels are generally purchased in the heat-treated condition.

In addition to the proper treatment prior to nitriding, another precaution is necessary. Unfortunately, the nitriding steels are susceptible to decarburization, so it is important that all decarburized material should be removed from the surface. It is recommended that, to have finished parts free from stress and decarburization, the best practice following heat-treatment and preceding nitriding is to rough-machine at least enough stock to ensure complete removal of the decarburized layer; redraw at a temperature at least as high as the nitriding temperature, to remove the stresses, and then finish-machine or grind.

Growth on Nitriding.—It is found that a slight growth takes place in the size of an article during the nitriding process. This growth is dependent upon the depth of the case and upon the temperature and time of nitriding. Fortunately, the growth is constant under given nitriding conditions, so that after it has once been determined for a given material or article it can be allowed for in machining before nitriding. It has been shown that an increase in diameter of 0.002 in. in a specimen having a case depth of about 0.030, is a fair example of the extent of the growth.

Protection against Nitirding.—It is frequently necessary to protect certain portions of the surface against nitriding. This is accomplished by either nickel plating to a depth of 0.0005 in. or more, or by coating with tin or tin solder. When the tin coating is used, the layer should be as thin as possible. Although the tin or lead-tin alloy has a melting temperature lower than that used in nitriding, the thin layer held to the work by surface tension will still afford protection. A thin coating is desired so that any excess that might otherwise be present will not drop on an article or a surface where protection is not desired. Other materials, such as glycerine and tin, shellac and tin oxide, or sodium silicate and chrome ore, are also used but they do not always ensure complete protection.

The final selection of a nitriding steel, therefore, involves consideration of the properties it is possible to develop both in the case and in the core, as well as the requirements of the particular service conditions.

Nitriding Equipment.—At the present time, furnaces are being built especially suited for nitriding. The essential requirement of such a furnace is that it must maintain a uniform temperature.

Electric-, gas-, and oil-fired furnaces are being used in commercial practice. It is obvious that pyrometric and automatic control equipment is desirable.

There are two general methods of nitriding—a batch, or intermittent, type of process and a continuous type. We will consider only the process of the first type, as it is being used almost exclusively at the present time. The typical batch-type furnace and the nitriding equipment used are shown in Fig. 4-XV. In this process the work to be hardened is stacked in a nitriding box in such a manner that the gas will circulate freely around it. No foreign matter—*i.e.*, no such thing as brick or asbestos—should be permitted in the box, because an undesirable catalytic effect may be induced by the presence of such material. The parts should be separated by nickel gauze, as nickel is not seriously affected by nitriding conditions, it being inert to the action of the hot NH_3 gases. The boxes should be of a material that will not disintegrate in the presence of ammonia gas. A number of different types of material have been used in the construction of the containers. Enameled low carbon steel containers have proved very successful, as the enamel is inert to the gases. One outstanding objection to this type of container is its tendency to chip or spall under service conditions. Almost all the other types of containers used are made from alloy steels of high chromium, high nickel content. For double effectiveness the containers are frequently made of Ni-Cr steel and enameled.

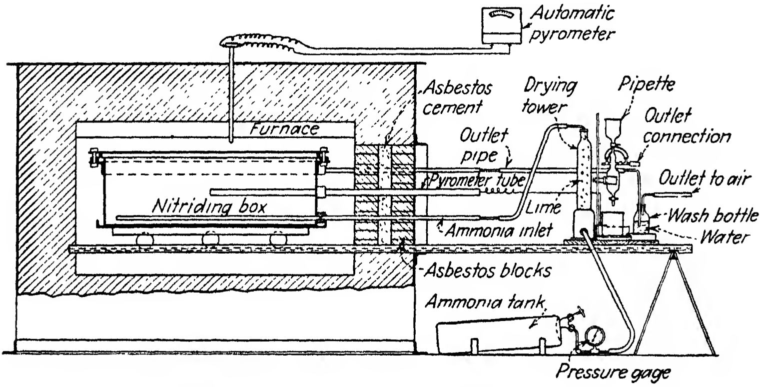

FIG. 4-XV.—Nitriding furnace and equipment. (*Courtesy of V. O. Homerberg.*)

The box must be designed so as to permit the lid to be sealed from the outside atmosphere.

The inlet and outlet tubes, which are made of nickel, should be so placed in the box that there will be uniform distribution of the active nitriding gas. The size of the box will vary with respect to the work on hand. The packed box is charged into the furnace and the ammonia gas, controlled by valves and gauges suitable for use with such gas, which is formed as the pressure is released, is passed through the box for the given time. The exact time at the nitriding temperature will depend upon the depth and character of the case required. The general temperature is 935 to 1000°F., and the time may range up to 90 hr.

Theory of Nitriding.—Exactly why a nitrided case is hard is not known. It is generally agreed that the ammonia gas coming into contact with the surface of the steel is decomposed or dissociated into nitrogen and hydrogen according to the reaction

$$2NH_3 + \Delta = 2N + 3H_2$$

At the moment of their liberation, both the gases are in the nascent state and while in this condition the nitrogen reacts with the alloying elements and the iron to form complex nitrides. The nitrides, which are in solid solution or in a fine state of dispersion in the case, diffuse into the steel and impart a hardness, which gradually decreases inwardly until it corresponds with the core. The rate of diffusion is a function of time, temperature, and, to some extent, the composition of the steel. The disturbing factor in this operation is that hydrogen, as well as the nitrogen, is liberated. It is not definitely known just what action the hydrogen may have, but evidence available indicates that it does retard and impair the process, thus being wasteful of ammonia.

The ammonia gas during the nitriding operation is partially broken up into hydrogen and nitrogen. The nitrogen is, of course, accessible to the steel, but only a portion of what is present alloys with the steel. The gas leaving the nitriding chamber will, therefore, consist of undissociated ammonia, nitrogen, and hydrogen. The ammonia gas is readily dissolved by water, while hydrogen and nitrogen are not appreciably soluble. Advantage is taken of this fact in the determination of the extent to which the ammonia has dissociated in the process.

The dissociation may be measured by the use of a suitable pipette apparatus, which is nothing more than a graduated glass chamber with suitable inlet and exhaust valves for trapping the gas and for admitting the water after the graduated chamber has been filled with the gas.

Control of the Process.—Temperature control is important in whichever process may be used. In the case of box nitriding, the time is computed from the time the box has reached the operating temperature. Temperature control is effected by use of thermocouples in the box or in the furnace, or in both. Homerberg and Walstead[1] state that increasing the nitriding temperature from 800 to 1400°F. causes increase in case depth with increase of temperature up to 1300°F., but drops off at 1400°F., while the hardness decreases as the nitriding temperature increases.

Many duplex cycles for nitriding have been suggested in an effort to reduce the time required for the production of a given case. In general, the cycle consists of nitriding first at a low temperature and then at a higher temperature. It is said that low-temperature nitriding produces maximum hardness and high-temperature processing, maximum penetration. Sergeson[2] recommends nitriding at a temperature of 950 to 1000°F., followed by a temperature of 1150 to 1200°F. Harsch and Muller[3] recommend a so-called *program-control* method, whereby the nitriding temperature is progressively changed from a low to a high temperature.

Structure.—Figure 5-XV shows a typical nitrided structure after nitriding at 960°F. for 48 hr. In general, the actual microstructure of the case and of the core are the same (sorbitic), although certain portions of the case etch darker than the core. The hardness of the case, as has been mentioned, is due to the precipitated nitrides, but these cannot be seen under the microscope.

The white layer, which can be seen at the surface of the nitrided specimen, is a nitrided layer consisting of a nitride of iron or a

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[1] HOMERBERG and WALSTEAD, A Study of the Nitriding Process, *Trans. A.S.S.T.*, Nitriding Symposium, **16**, 67–110 (1929).

[2] SERGESON, R., Investigations in Nitriding, Nitriding Symposium, *Trans. A.S.S.T.*, **16**, 145–174 (1929).

[3] HARSCH and MULLER, Deep Nitriding by Program Control, *Metal Progress*, **21** (No. 1), 74–76 (1932).

complex nitride of iron and one or more of the alloying elements. It is this very thin layer, which is generally from 0.001 to 0.002 in. thick, that provides maximum resistance to corrosion.

Physical Properties of Nitrided Steels. *Hardness versus Depth.* Since the nitrided steels are in direct competition with carburized steels for application and service under certain working conditions, a comparison of the depth-hardness figures will serve to illustrate why the nitrided parts are superior in many applications.

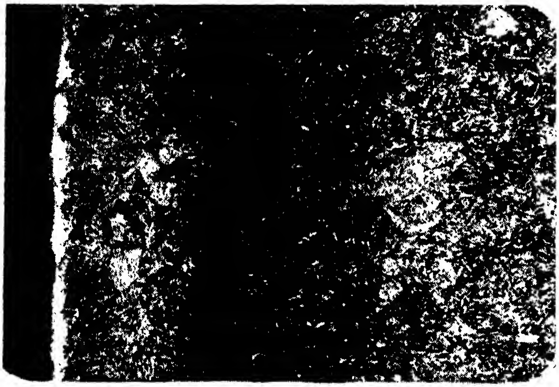

FIG. 5-XV.—Typical nitrided structure. Nitrided at 960°F. for 48 hr. × 100. Nital etched. (*Courtesy of V. O. Homerberg.*)

Figure 6-XV, from Sergeson's investigation,[1] shows the relation between the depth of penetration and Vickers hardness (modification of Brinell hardness) of a chromium-aluminum-molybdenum steel nitrided 90 hr. at 975°F., and a chromium-vanadium-nickel-molybdenum steel, which have been carburized 8 hr. at 1650°F., hardened by double quenching, and tempered at 300°F. It can easily be seen that the nitrided case with a surface hardness of 1100 Brinell is far superior to the 850 Brinell hardness of the carburized steel up to a case depth of 0.030 in., below which the carburized steel is harder. The higher surface hardness of the nitrided steel, which is maintained to a sufficient depth, means increased resistance to wear.

A number of investigations have been made to determine the effect of heating to different temperatures, on the room-tempera-

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[1] SERGESON, R., Investigations in Nitriding, Nitriding Symposium, *Trans. A.S.S.T.*, **16**, 145–174 (1929).

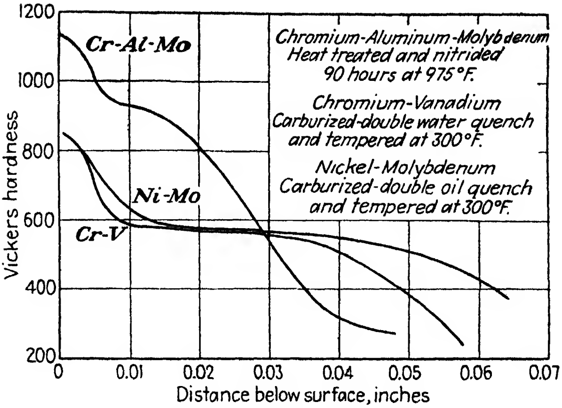

FIG. 6-XV.—Relation of hardness penetration of nitrided chromium-aluminum-molybdenum, carburized nickel molybdenum, and chrome vanadium steels. (*After Sergeson.*)

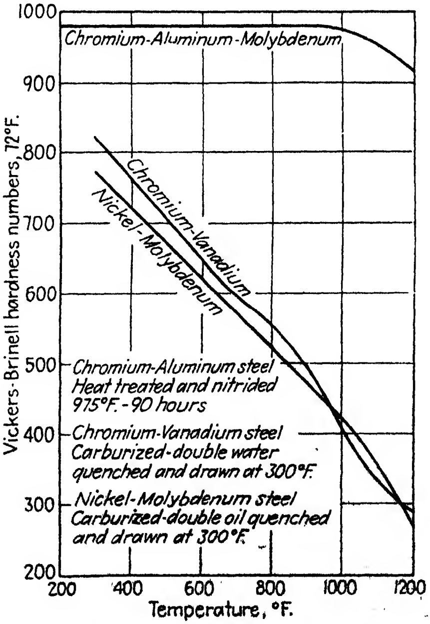

FIG. 7-XV.—Effect of reheating temperature on hardness of surface of nitrided and carburized steels. (*After Sergeson.*)

ture hardness of the nitrided and carburized cases. It has been found that the nitrided case, when heated to around 1000°F., suffers very little or no loss in hardness, while the carburized case falls off in hardness as the temperature is progressively increased. In Fig. 7-XV this effect is shown for the steels discussed above.

Abrasion and Wear.—All investigations have shown that nitrided steels have high resistance to wear. Malcolm,[1] in making a comparison between a nitrided and a stainless steel, found that the stainless was badly abraded after 1000 revolutions under 200 lb. pressure, while the nitrided steel under the same conditions withstood 100,000 revolutions with comparatively little wear. Guillet,[2] in his investigation on the wear resistance of nitrided and cast-iron automobile cylinders, found that after 18,630 miles of travel there was 0.016 in. wear on the cast-iron cylinders while on the nitrided cylinders the wear was only 0.0008 in. In aviation motors using heat-treated cylinders the wear was 0.003 to 0.004 in., while the wear on the nitrided cylinders was not measurable.

Uses of Nitrided Steels.—Harder[3] has given an excellent résumé on the application of nitrided steels, from which we shall quote.

> The use of nitrided materials in general is dependent upon a high degree of hardness and resistance to wear, abrasion and erosion; to some extent upon resistance to corrosion; and possibly to a high endurance limit when subjected to repeated bending stresses under the previous service conditions.

The most important uses for nitrided steels in the automobile industry are those for "cams, camshafts, crankshafts, piston pins, pump shafts, push rods, seats for valves, steering worms, taper gears, tappets, timing gears, valves, valve pusher-rod rollers, valve stem guides."

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[1] Malcolm, V. T., Use of Nitrided Steel in High-Temperature—High Pressure Steam Service, Nitriding Symposium, *Trans. A.S.S.T.*, **16**, 205–222 (1929).

[2] Harder, O. E., Nitriding for the Engineer, *Metals and Alloys*, **2**, 139 (1931).

[3] Harder, O. E., Nitriding for the Engineer, *Metals and Alloys*, **2**, 132–142 (1939).

There are many other applications in airplane motors, in Diesel engines, and in many other miscellaneous parts.

Suggested Questions for Study and Class Discussion

1. What is the object of casehardening? What are the requirements and features of this process?

2. What grades of steel are used in carburizing?

3. In the carburizing process, how is the carbon absorbed into the steel and what determines the amount absorbed?

4. What media might be used to bring the carbon to the surface of the steel?

5. In the heat-treatment of carburized material what type of material, structurally, must we consider as being present?

6. Outline a heat-treatment that will produce a fine grain structure throughout the case and core of steel parts that have been carburized. Treatment other than that used with grain-controlled steels.

7. What advantages are gained by using grain-controlled steels in carburized steel parts?

8. Describe the cyaniding process of heat-treatment.

9. In what way does the case produced by cyaniding differ from that produced by carburizing?

10. What are the advantages of the cyaniding method for casehardening?

11. Why are special steels used in the nitride process for casehardening?

12. Describe the nitriding process.

13. What differences exist in the heat-treating of carburizing steels and nitrided steels?

14. In what applications would nitrided steels displace carburized steels? Why?

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CHAPTER XVI

ALLOY STEELS

In the previous discussion of carbon steels and their heat-treatment, frequent reference has been made to the fact that certain alloys, if added to steel in given amounts, will impart to the steel properties that are definitely superior to those developed in straight carbon steels. The commercial plain carbon steels are not strictly iron and carbon alloys but contain, besides, other elements, such as manganese, phosphorus, sulfur, and silicon, which have their influence upon the properties of the finished steel. An alloy steel should, then, be described as one whose physical properties have been considerably improved by the addition of some element other than iron and carbon in quantities larger than those ordinarily found in the plain carbon steels.

In considering steels of special composition, it is desirable to differentiate between the so-called *low alloy, structural,* or *constructional* steels and "special" steels. The former terms are used in reference to steels to which a certain amount of special element has been added in order to improve the physical properties that are, in most cases, mechanical. *Special* steels are, properly, those steels that possess certain definite unusual properties that are not ordinarily found in plain carbon steels. To differentiate the latter from the former class, there may be cited such examples as high-speed tungsten tool steel, rustless chromium steels, and others.

It should be obvious that two, three, or even four special elements may be used, with a view toward imparting the beneficial influence of each. Alloy steels that contain one special element are called *ternary* steels and are considered to be made up of three constituents; *i.e.*, iron, carbon, and the special element. Those that contain two special elements are called *quaternary* steels, because of the presence of four elements: iron, carbon, and the two special elements.

A study of such mixtures involves the study of a complex system involving, in the case of a ternary alloy, a three-component relationship. To represent this relationship graphically four axes are required, the variables being iron, carbon, special element, and temperature. For the purpose of accurate study the best procedure is actually to construct a triangular space model, as a printed description would be difficult to study and interpret. The difficulty of a graphical representation of the quaternary steels is plainly apparent.

One method used to eliminate this difficulty in the study of such alloys is to select a certain composition of alloying element and construct a cross-sectional diagram for this composition. This gives a quasi-binary diagram in which iron plus a given per cent of alloying element with varying per cent carbon is one component and temperature, the other.

It must be stressed that in the study of the cross-sectional diagrams or constitutional diagrams of a ternary system there is an essential difference between these diagrams and the equilibrium diagrams of the binary alloy systems. This latter system contains all the important information, namely, temperatures of various transformations, the composition of the phases, and their relative amounts; while the cross-sectional representation gives only the temperature of phase changes. The composition and amounts of the various phases can be obtained only from the entire space model.

The terminology used to designate the constituents of the iron-carbon system is also used to describe the more complex systems. Thus, we will have delta iron, gamma iron, alpha iron, cementite, marteniste, troostite, sorbite, and pearlite, as before; but it must be remembered that the composition of the constituents will be different. For example, if one of the special elements is soluble in austenite, we will have, instead of a solid solution of carbon in gamma iron, a solid solution of carbon, gamma iron, and special element. For proper differentiation between constituents of carbon and alloy steels, which are similar in nature but chemically different, we may use the terms "chromium-austenite," "nickel-ferrite," "iron-tungsten carbide," etc., depending upon the alloying element present.

SPECIAL ELEMENTS

The special elements used in the production of alloy steels include chromium, tungsten, molybdenum, vanadium, manganese, nickel, silicon, copper, titanium, cobalt, aluminum, and zirconium.

These special elements can be divided into two groups: those that form carbides, either simple carbides resulting from chemical combination of the special element and carbon, or more complex carbides of the special element and iron; and those that do not combine with the carbon to form carbides. Those elements that, on slow cooling, form carbides are manganese, chromium, tungsten, vanadium, titanium, columbium, and molybdenum; while such elements as nickel, copper, and silicon, which on slow cooling remain soluble in gamma iron and alpha iron, do not form carbides, but form solid solutions. The carbides formed by the above elements may be wholly or partially dissolved in gamma iron at a high temperature, but they frequently separate as free carbides on slow cooling.

INFLUENCE OF SPECIAL ELEMENTS UPON THE EQUILIBRIUM DIAGRAM

In the successful heat-treatment of plain carbon steels, the importance of a knowledge of the position of the critical temperatures, the Ac_3 on heating and the Ar_1 on cooling, needs no emphasis. If to these plain carbon steels certain additions of special elements are made, above the amounts normally present, the positions of the critical temperatures are shifted considerably. It is found that with such additions the critical temperatures may be either raised or lowered, depending upon the element added. It is obvious, therefore, that with such changes in the critical temperatures the procedures involved in heat-treatment also will be changed. Prior to a study of the heat-treatment of alloy steels it becomes necessary, in view of the changes produced in the critical temperatures, to study the general effects of (1) the addition of a special element to iron and (2) the addition of a special element to the binary iron-carbon alloys.

Why do certain elements raise the allotropic transformation and others lower it? In answering this question, let us consider the space lattice structure of the different metals. First, let us recall a few facts about iron. Gamma iron, which exists from the A_3 to the A_4 point, is face-centered, and alpha iron, which exists below the A_3 point, is body-centered. In general, it is found that metals having the same space-lattice are more soluble in each other. From this it should follow that the metals that are body-centered should be more soluble in alpha iron and those with face-centered lattice should be more soluble in gamma iron. It should be reasonable to conceive, then, that face-centered metals dissolved in gamma iron should tend to stabilize, oppose, and retard the transformation of gamma iron into a lattice in which they are less soluble; *i.e.*, the rate of transformation would be retarded and thus tend to lower the critical transformation temperature. On heating, the metal will tend to adjust itself more rapidly to the lattice in which it is most soluble and should, therefore, hasten the alpha-gamma transformation and tend to lower the upper critical transformation temperature. Similarly, the body-centered metals, being more soluble in alpha iron, should tend to cause the allotropic transformations to occur at a higher temperature. It is found that the body-centered cubic metals, chromium, tungsten, molybdenum, and vanadium, raise the A_3 point, while the face-centered cubic metals, such as nickel and copper, lower that point. Several of the elements, however, are found to be out of line with these facts. Aluminum, which is face-centered, should lower the critical point but it is reported as raising it. Silicon, which has a cubic lattice (diamond arrangement of the atoms), raises the A_3 point, from which we can assume that such a structure is more readily soluble in a metal having a body-centered lattice. Manganese, which lowers the A_3 point, is said to have a tetragonal face-centered lattice. From these facts it must act like face-centered cubic metals, which are more soluble in gamma iron. Carbon is also out of line, in that it tends to lower the A_3 transformation, whereas, since it is reported as having the same lattice as silicon, it should raise it, instead.

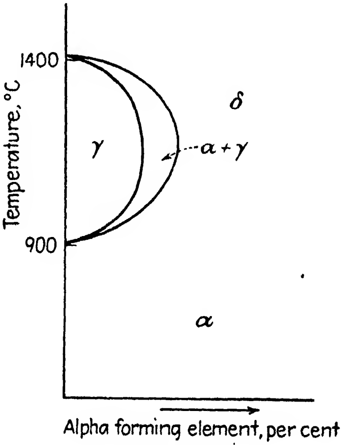

FIG. 1-XVI.

The typical diagram (Fig. 1-XVI) illustrates the effect of those body-centered elements that raise the A_3 point. It is also found that the elements that raise the A_3 point generally lower the A_4 point; *i.e.*, they delay the transformation of body-centered delta iron into face-centered gamma iron. From this, then, it should be found that, with increasing amounts of dissolved body-centered element, the A_4 should be lowered and the A_3 should be raised until a percentage is obtained where the two points should merge. Figure 1-XVI shows the result of such a merging at a point to form what is known as the *gamma loop.* Obviously, then, if the special element is present in excess of this point, the solvent iron will not undergo an allotropic transformation but will solidify as body-centered delta iron and remain body-centered to room temperature, although it is now called alpha iron. The two-phase region denotes the sluggishness toward transformation and shows that the body-centered phase (delta and alpha both being body-centered) can and does exist with the face-centered gamma phase over a range of temperatures.

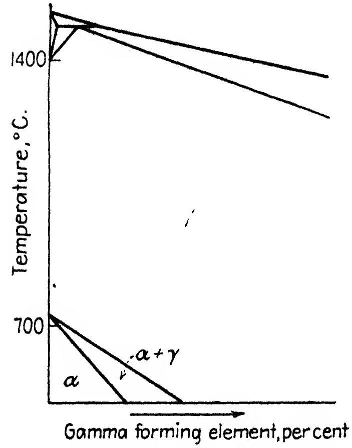

Fig. 2-XVI.—Typical diagram for iron with gamma-forming (face-centered) alloy.

The typical diagram (Fig. 2-XVI) illustrates the effect of the face-centered gamma-forming elements that tend to lower the A_3 and raise the A_4 critical transformation. From this it can be seen that with increasing amounts of special element the A_3 critical point will progressively be lowered until a point is reached where the alloy transforms at extremely low temperatures. The solvent iron in this case will, under ordinary conditions, solidify as face-centered gamma iron and remain in that condition to room temperature.

Effect of Elements That Lower the Critical Range.—Now that the effect of adding certain elements to iron has been observed, it becomes essential to determine the changes that take place when a gamma-forming element is added in varying percentages to the iron-carbon alloys. To study what changes are brought about, it is only necessary in this case to study the critical-range portion of the equilibrium diagram.

Figure 3-XVI shows the effect on the iron-carbon critical range of making progressive additions of an element that lowers the critical range. The normal position of the critical range is shown in the lines *GS*, *SE*, *PSK*. Suppose that, upon the addition of 5 per cent of the alloy element, the critical range assumed the position shown by the lines *G'S'*, *S'E'*, *P'S'K'*. It is immediately apparent that the entire critical range has been lowered, the A_3 and the A_{cm} lines being shifted so that the eutectoid point has moved to the left. Thus, a slowly cooled bar of this steel containing 0.5 per cent carbon will no longer be composed of ferrite and pearlite but will be entirely pearlitic; in other words, the pearlite of eutectoid composition no longer contains 0.90 per cent carbon but now contains 0.50 per cent carbon. The physical properties of this steel will consequently approach those of 0.90 per cent carbon unalloyed steel. As a result of these shifts, it would be theoretically possible to quench from lower temperatures for hardening than in the case of the unalloyed carbon steel, while actually, owing to sluggishness, these low temperatures are not practical. Another effect will be produced, which is not so apparent; namely, that it will not be

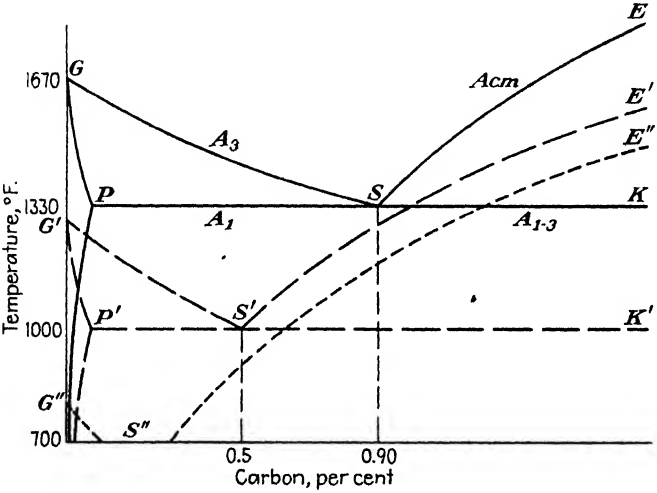

FIG. 3-XVI.—Effects of progressive additions of an alloy element which lowers the critical range and decreases the percentage of carbon in the eutectoid. (*After A. A. Bates.*)

so essential to cool the steel so rapidly in order fully to harden it or make it martensitic.

If sufficient alloy addition is made to the steel, the effective cooling rate may be so retarded that it is possible to harden fully even on air cooling. If such conditions do exist, the steel is known as an *air-hardening* steel. In a similar fashion the terms "oil-hardening" and "water-hardening" steel are derived.

With further increase of alloy elements of this type, the critical range will continue to assume lower positions on the temperature scale and may eventually reach or go below room temperature. Should the percentage of alloy element be sufficient to lower the range below room temperature, the steel will remain untransformed, or austenitic, at atmospheric temperatures. Such alloys have peculiar physical properties, which do not follow the same relationship to carbon content as do the ordinary iron-carbon steels. Their physical properties are dependent primarily upon the grain size of the austenite and the extent to which carbide has precipitated from the austenite.

In order that some of the peculiarities of such alloy steels may be understood, let us consider a bar 1 in. in diameter that contains 0.90 per cent carbon, and whose critical range, indicated in Fig. 3-XVI, is $G''S''E''$. The eutectoid temperature, in this case, is below 700°F., possibly occurring somewhere in the vicinity of 600°F. Upon being heated to, say, 1700°F., the bar will be completely austenitic. When this alloy is cooled at a moderate rate from this temperature, it will first encounter the A_{cm} line ($S''E''$) at about 1100°F., and carbide will precipitate at the grain boundaries and in the cleavage planes. Upon further cooling, the temperature will reach the vicinity of 600°F., where the A_{1-3} critical is encountered, but, instead of the usual transformation to pearlite at this critical temperature, the steel transforms to hard martensite. This is due to the lack of sufficient atomic mobility at this low temperature to allow the transformation to occur. If this same bar is reheated to 1700°F. and then water-quenched, neither carbide nor martensite will be formed; instead, the soft, tough, and ductile austenite will be retained.

One other point of importance should be mentioned about these low critical-range types of steel of which the manganese and nickel steels of high percentages are the most important. This is a peculiarity in regard to the effect of nickel and manganese on the *Ar* and *Ac* points; that is, for a given carbon composition and percentage of alloy, the *Ar* point is lowered much more effectively than the *Ac* point. The distance between the A_1 transformation points is even quite apparent on slow heating and cooling. This makes it most essential that the *Ar* and *Ac* points should be known for successful heat-treating operation.

Effects of Elements That Raise the Critical Range.—The majority of the common steel alloying elements raise the critical range. These elements may be divided into two classes:

1. Those elements that shift the eutectoid point to the right. These include principally aluminum, cobalt, and copper.

2. Those that shift the eutectoid point to the left. The most important of this class are chromium, tungsten, vanadium, molybdenum, and silicon.

Elements of the first class are used only in moderate amounts or for very special purposes. This group will not, therefore, be considered separately. Those of the second class, however, are all elements used quite extensively. The effect of the five elements mentioned is similar as regards the critical range. These effects are illustrated in a somewhat simplified form by Fig. 4-XVI, which represents the general effect of continually increasing additions of these elements.

The critical range of plain carbon steel is represented by the lines *GS*, *SE*, *GP*, *PN*, *PSK*. When an addition of several per cent of the alloy element is made, the eutectoid point *S* is moved upward and to the left, to the position S_1. The point *P*, which denotes the maximum solubility of carbon in alpha iron, has been moved upward and to the right, to the point P_1. Upon further additions, this same action will continue so that the various lines will reach some position as denoted by G_2S_2, S_2E_2, G_2P_2, $P_2S_2K_2$. At this point, however, a new effect makes its appearance; two new lines, H_2J_2 and I_2J_2, which end in the point J_2, begin to approach G_2 rather closely from high temperatures. This intrusion may be readily understood if we refer to the upper left-hand corner of the iron-carbon equilibrium diagram, in which there are several related triangular areas bounded by the lines *HN* and *JN*. In the triangle *HNJ* there are two forms of iron in equilibrium with each other; *i.e.*, delta, which is identical with alpha iron, and gamma. (In Fig. 4-XVI, H_2J_2 and I_2J_2 are the same as *HN* and *JN* in Fig. 2-VI.)

The importance of this area, which was neglected for the most part in the study of the iron-carbon alloys, becomes manifest when increasing amounts of given elements are added that raise the critical range. When such elements as chromium, tungsten, molybdenum, and vanadium are added to the iron-carbon alloys, the effect is to raise the lower critical and to lower and widen the range in which delta iron exists. Thus, with increasing additions of alloy element, the points *J* and *G* merge, and eventually the lines *HJ* and *GP* become one line (H_3P_3) and move further toward the right. The lines I_2J_2 and G_2S_2 will likewise become a single line I_3S_3.

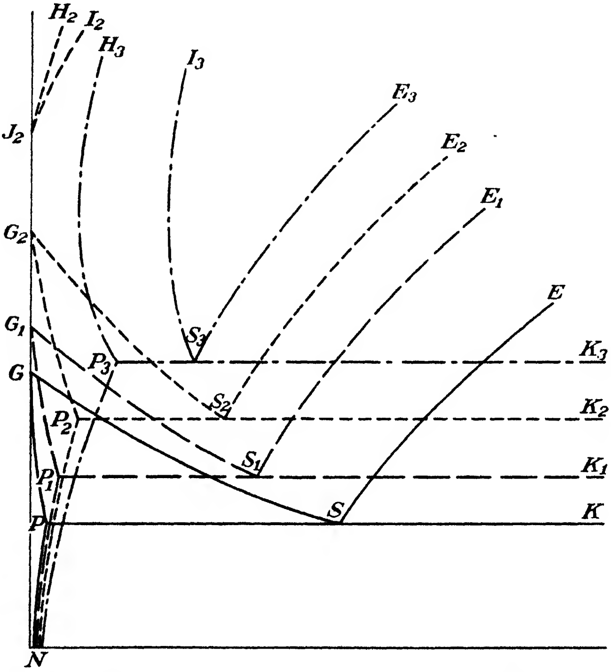

FIG. 4-XVI.—Effects of progressive additions of an alloy element which raises the critical range and decreases the percentage of carbon in the eutectoid. (*After A. A. Bates.*)

Even though the diagram has become somewhat distorted, it need cause no trouble in the consideration of the heat-treatment of these alloys, as they will follow the same general rules of equilibrium interpretation that have been applied previously to the plain carbon steels. There are certain peculiarities, but these for the most part can be explained by the position of the critical lines. For further clarification let us study Fig. 5-XVI and, first of all, alloy X. We can see from this diagram that at any temperature below actual melting this alloy remains ferritic and, hence, cannot under any circumstances be heat-treated. It will, however, undergo the usual process of recrystallization when cold-worked and subsequently annealed at a high temperature. Thus, hot work and cold work may be performed on it, if we consider hot work as the deformation carried out above the recrystallization temperature.

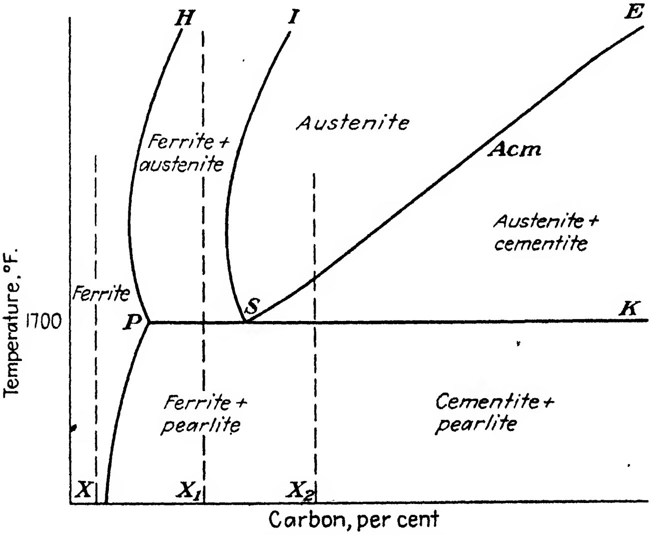

Fig. 5-XVI.—Critical range of an alloy steel in which delta and alpha areas have merged. (*After A. A. Bates.*)

The alloy X_1, according to the diagram, can, on heating, be transformed partly to austenite and can, therefore, be heat-treated by heating and cooling through the critical range. The alloy X_2 is also heat-treatable, in that by heating above the line SK it transforms in part to austenite.

Considering the case of those alloys whose critical range is elevated as in the line G_2S_2, S_2E_2, $P_2S_2K_2$ of Fig. 4-XVI, we find that this series is all heat-treatable in the usual way. The full-annealing and quenching temperatures will, as can be seen from the position of the critical range, be higher than normal. Due account must also be given to the effect of the lowering of the carbon percentage in the pearlite, together with the increased carbon in the ferrite.

At this point we encounter an unexpected irregularity in the expected reaction of this group of steels. When we considered the manganese and nickel steels, it was pointed out that they can be hardened fully, even upon comparatively slow cooling. By inference, we should expect those alloy steels that exhibit a high critical range to require a very rapid rate of cooling for full hardening. This, however, is usually not the case, in that it is found that they also may be hardened by comparatively slow cooling.

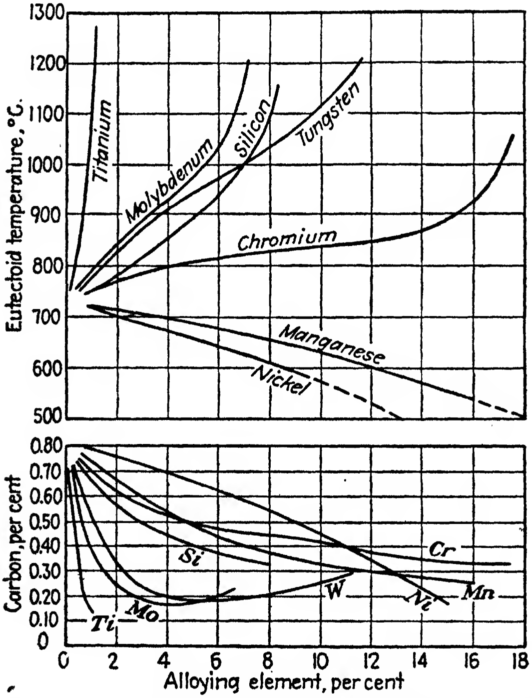

FIG. 6-XVI.—Eutectoid composition and eutectoid temperature as influenced by several alloying elements. (*After E. C. Bain.*)

Eutectoid Composition and Temperature in Alloy Systems.—In the previous discussion, the general effect of progressive additions of an alloy element on the critical range and the percentage of carbon in the eutectoid has been shown. In Fig. 6-XVI[1] are shown (1) the actual eutectoid temperature when a given percentage of alloying element is present, and (2) the proportion of carbon that, with a given alloy content, will result in the minimum temperature for complete austenite formation, *i.e.*, the eutectoid composition. The effect of each of the elements on the eutectoid temperature and composition is shown to be individual and characteristic. Some of the elements are found to lower the temperature of austenite formation, while others raise it. However, none of the elements are found to shift the eutectoid composition to higher carbon values.

The above series of compositions all correspond to binary, or two-component, alloy steels. In the case of steels that have two or more alloying elements, there are only a few cases in which the systems have been studied over a range of carbon contents. The outstanding example is the 18 chromium, 8 nickel stainless, which will be discussed subsequently.

INFLUENCE OF ALLOYING ELEMENTS

It must be noted that this inquiry into the manner in which alloying elements affect the properties of steels pertains to that group of steels in which the alloying element is present in amounts not exceeding 4 or 5 per cent. It is also assumed that these alloys contain enough carbon to make them responsive to heat-treatment. The special alloys that contain very large proportions of the alloying elements added for special purposes, such as stainless steel, Hadfield's manganese steel, high-speed steels, etc., will be discussed in more detail under a separate section.

It has previously been learned that the iron-carbon alloys are affected by heat-treatment and undergo definite reactions that are responsible for the valuable properties thus produced. However, certain particular physical properties are not attained by this class of steels but are best met by the addition of small amounts of some alloying elements to the steel. Our purpose in this section of the text (Alloy Steels) is to make inquiry into the way in which the alloying elements enhance the properties of steel.

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[1] **Bain, E. C., "Functions of the Alloying Elements in Steel," A.S.M., Cleveland, Ohio, 1939.**

Reaction Rate.—The first important thing—one that should be kept firmly in mind—is that the rate of the reaction, *e.g.*, austenite → ferrite and carbide, is the principal factor in determining the distribution; *i.e.*, whether it will be fine or coarse dispersion of the two principal constituents of steels, ferrite and carbide. This distribution is, in turn, a large factor in the control of the properties of the steel. The tendency of alloy steels, under similar conditions of treatment, is to result in a much finer eutectoid structure. This is frequently termed the *sluggishness* of the steel.

We have seen in the development of carbon steels how the rate of cooling affects the rate of the above reaction and we have also learned that, with the introduction of alloying elements, the reaction rate will be retarded. This means that with the same rate of cooling, with two steels of the same carbon content—the one containing an alloying element and the other no alloying element—the one containing the alloying element will have a retarded reaction rate, the transformation will occur at a lower temperature, and the product formed will be considerably harder. In brief, this means that the introduction of an alloying element into the austenite solid solution will decrease the rate of the reaction so that greater undercooling may occur, thus bringing the austenite into a low-temperature range for transformation. The reaction at this point of undercooling is rapid and the product will consist of a finer dispersion of metallographic constituents of higher hardness. With more and more rapid cooling this product, ferrite and carbide, becomes finer and finer until the two phases can no longer be separated by the highest resolution of the microscope. Beyond this particular rate of cooling a different reaction takes place at a much lower temperature and a different reaction product is produced, namely, martensite. The metal has, in this case, been cooled at the critical quenching rate and has escaped the austenite → pearlite reaction.

Depth of Hardening.—Since alloy additions, except aluminum and cobalt, decrease the critical quenching rate necessary for hardening, it follows that, for the same section size, a steel containing an alloy will harden deeper than a corresponding straight carbon-steel section quenched into the same medium. Figure 7-XVI shows the hardness distribution across the same section of three steels identically quenched. Thus, it can be seen that the increased alloy content of the fully hardened section accomplishes the necessary decrease of the critical quenching rate, so that even the metal in the center of the section that cools the slowest is still able to harden. Through this effect, which is probably the most important contribution of alloying elements, large sections may be fully hardened and be ready for subsequent tempering. Tempering, as we know, develops the spheroidal carbide distribution and leads to maximum toughness for any strength. It is found that practically all the elements, if they are dissolved at the temperature of heating, will contribute to deep hardenability.

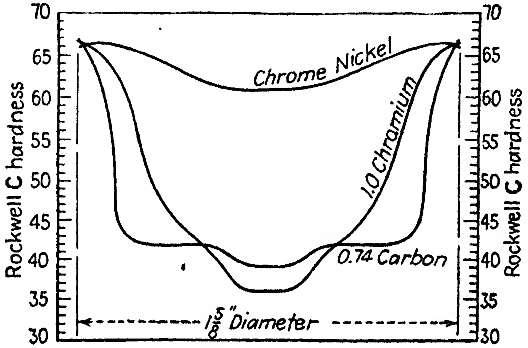

Fig. 7-XVI.—Deeper hardenability contributed by alloying elements. Shallow-carbon steel; medium-chromium steel; deep-chrome-nickel steel. (*After E. C. Bain.*)

The effect of alloying elements upon the reaction rates answers the unexpected irregularities in the effects of additions of nickel, which lowers the critical range, and chromium, which raises the critical range, yet which both produce steels that harden on slow cooling. The effects of adding moderate percentages of nickel and chromium are shown qualitatively in Fig. 8-XVI. Curve 1 shows the reaction-rate curve for a plain carbon steel of moderate grain size. Curve 2 shows the curve for a nickel steel. The effect, as can be seen, is to lower the critical range and reduce the rate of transformation. Curve 3 shows how chromium raises the critical range but yet retards the rate of transformation. Thus, the plain carbon steel must be cooled at a rate shown by the curve C_1a in order to harden fully, while the nickel steel and the chromium steel can be hardened by a slower rate, as is shown by the curves C_2b and C_3c, respectively.

The influence of individual alloying elements on the S-curve may be summarized in general terms as follows:[1]

Manganese.—The general effect of manganese is to retard the transformation at all temperature levels. Through a study of the transformation rates it has been possible to explain logically the well-known fact that manganese is an effective element for increasing the depth of hardening.

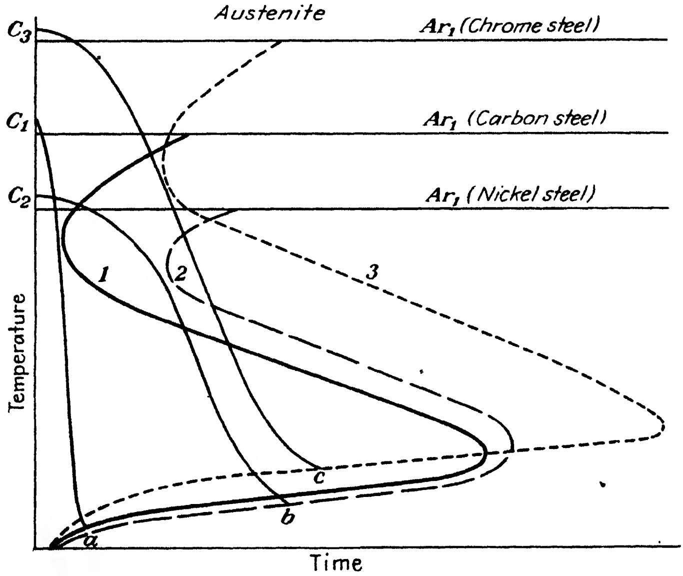

FIG. 8-XVI.—Comparative reaction rate curves for plain carbon steel, nickel steel, and chromium steel. (*After A. A. Bates.*)

Nickel.—Nickel, like manganese, is an element that consistently retards the transformation at all temperature levels, but much less markedly than manganese.

Chromium.—The influence of this element in the lower carbon ranges is found to be very effective in retarding the transformation in the Ar_1 range. This indicates the reason why many of the chromium-bearing steels have marked air-hardening tendencies. The retardation effect of chromium on the Ar_1

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[1] DAVENPORT, E. S., Isothermal Transformation in Steels, *Trans. A.S.M.*, **27**, 837–886, (1939).

appears to be greater than that of manganese, which in itself is more effective than either carbon or nickel.

Molybdenum.—This element in percentages up to about 0.5 per cent appears to be much more effective than chromium in retarding the transformation at temperatures above 1000°F. It has been found that retardation of the transformation increases with increased molybdenum content.

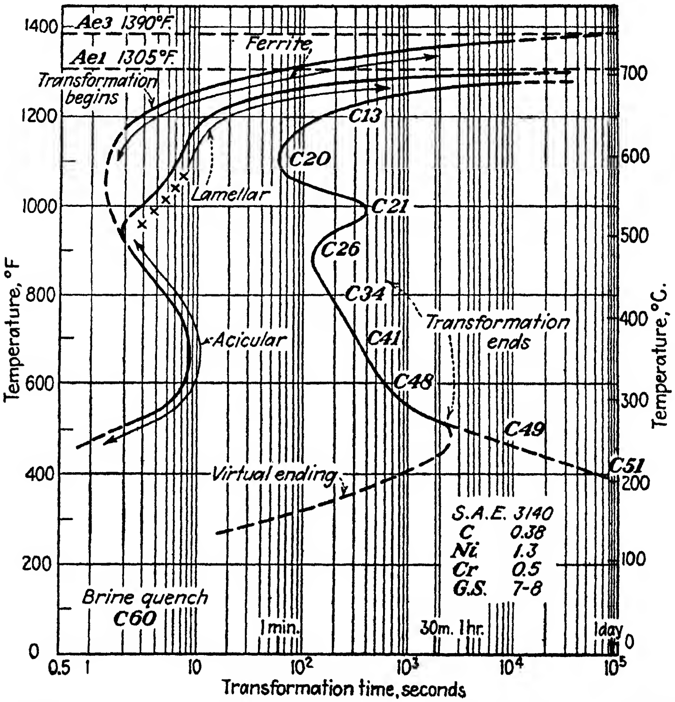

FIG. 9-XVI.—Isothermal transformation curve for S.A.E. 3140. (*Courtesy of E. S. Davenport.*)

Vanadium.—The influence of small amounts of vanadium (0.20 to 0.30 per cent), with the vanadium actually in solid solution, shows that it retards the transformation at and below the Ar_1 range, but at higher temperature levels it appears to accelerate it.

Cobalt.—Cobalt, which has been found to be an element that promotes shallow hardening, actually accelerates the transformation of austenite in the Ar_1 range. There is no ready explanation for this unique influence. However, it has been suggested that it may be related to the fact that cobalt undergoes an allotropic transformation not far from the temperature range in which iron transforms.

Davenport[1] has reported the *S*-curves for a number of the S.A.E. alloy steels, of which two examples, Fig. 9-XVI and Fig. 10-XVI (S.A.E. 3140 and S.A.E. 4340), are reproduced, the further to illustrate the effect of the alloying elements on the isothermal transformation.

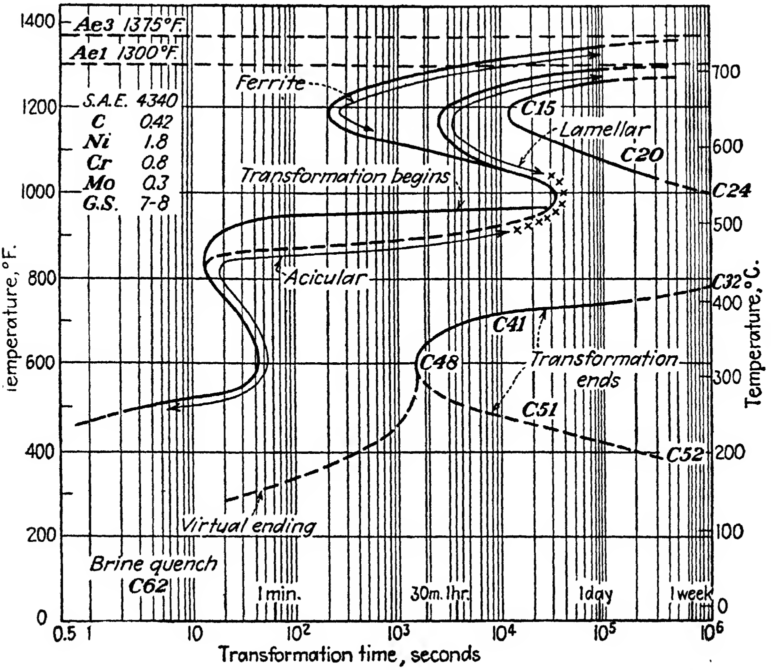

FIG. 10-XVI.—Isothermal transformation curve for S.A.E. 4340. (*Courtesy of E. S. Davenport.*)

Grain Size.—The effect of alloying elements on grain size has been discussed under the chapter on grain size, but it may be well briefly to review the effects. Plain carbon steels can be made moderately deep hardening by means of coarsening the austenite

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[1] DAVENPORT, E. S., Isothermal Transformations in Steels, *Trans. A.S.M.*, **27**, 837–886 (1939).

grain size, but the results of this practice are not too desirable since the product of the coarsened grain size will have low ductility and low impact strength. It is desirable, therefore, to retain the fine-grained structure and induce the necessary hardenability by means of alloy additions. This not only denotes that the sections have the capacity to form fully hard martensite by quenching, but also implies that, should the section size be extremely large or the quenching rate not rapid enough to form martensite throughout, the products formed will be harder and stronger than those that would be formed under similar circumstances with plain carbon steels.

It is found also that the carbide-forming elements, owing to their stability and reluctance toward solution, offer great resistance to grain growth. Therefore, by proper heat-treatment, these steels may contribute both deep-hardening characteristics and fine-grain austenite. This will reflect not only in the hardenability of a given composition but also in the toughness of its harder structures. A similar combination may be secured by using an element, such as manganese, that will contribute to the hardenability, together with a deoxidation practice which will ensure fine grain.

Stabilization on Tempering.—Certain of the elements when added to steels form carbide particles, which during tempering will coalesce very slowly. The softening rate, in this case, is controlled by the rate at which the carbon diffuses through the ferrite matrix. Since these dispersions are reluctant to coalesce, they will retain their strength after much higher reheating temperatures than will carbon steels. This will be reflected in their creep-resisting properties, which depend upon the preservation of a fine dispersion of carbide particles, and the fact that these steels may be tempered to higher temperatures and yet be as strong as a carbon steel that has been tempered at a much lower temperature to obtain the comparative strength. The relief of internal stress in martensite depends, as we know, on the tempering temperature. From this it can be inferred that the alloy steel will have a lower internal stress than a carbon steel of the same hardness. One of the causes of low ductility and low impact strength is internal stress upon which the external stress of the test is imposed. Therefore, the higher possible tempering temperature of alloy steels, with its accompanying lower internal stress, results in greater toughness at comparable hardness.

The Function of Alloying Elements.—It is evident that the structure of the steel is a most important factor in the control of its properties. Let us briefly sum up the functions of the alloying elements.

1. They improve the physical properties in one of the following three ways:[1]
   - *a.* By changing the state of dispersion of the carbide in the ferrite.
   - *b.* By changing the properties of the ferrite.
   - *c.* By changing the properties of the carbide.

These effects should be evident when we stop to consider that, should the element be soluble in the ferrite, as are nickel, vanadium, etc., the properties will most certainly be different from the ferrite in plain carbon steels. Similarly, the properties of the carbide should be changed, and both effects together will change the properties of the eutectoid.

Moreover, the addition of an alloying element to a plain carbon steel shifts very considerably, at times, the positions of the critical points. This fact is important when we consider that the entire process of the heat-treatment of steels is based on the position of the so-called *critical points.* Thus, it is important to know the action of a given element so that the necessary changes in the temperature of heat-treatment can be made.

2. Through the above-mentioned functions, the alloy additions accomplish one or more of the following:
   - *a.* Increase the hardenability.
   - *b.* Retain fine-grained austenite.
   - *c.* Retard softening during tempering.
   - *d.* Strengthen the pure-ferrite matrix.

The alloying elements, when they are present in amounts exceeding the percentage limits of the above class, are what we have termed the *special* alloys. The effects of these additions and the characteristics produced will be discussed in a separate section.

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[1] Bain, E. C., Some Characteristics Common to Carbon and Alloy Steels, *Year Book Am. Iron Steel Inst.*, **24**, 86–128 (1934).

Summary of the Effects of Alloying Elements.[1]—The effects of the usual alloying elements are as follows:

- To Strengthen Ferrite
  - P, very strong
  - Si, Mn, strong
  - Ni, Cu, Cr, Mo, W, fairly strong
  - V, weak
- To Confer Depth Hardening
  - Mo, W, Mn, Cr, strong
  - Si, medium
  - Cu, Ni, weak
  - V, very weak
- To Restrain Grain Growth of Austenite
  - Al, very strong
  - V, Ti, Zr, fairly strong
  - Mo, medium
  - Cr, weak
  - Si, weak
  - Mn, very weak
  - Ni, practically no action.
- To Form Carbides
  - Ti, Cb, V, very strong
  - Cr, Mo, W, strong
  - Mn, fairly strong
  - P, Ni, Cu, Si, no carbides
- To Stabilize Martensite and Troostite on Tempering
  - Mo, W, strong
  - V, fairly strong
  - Cr, weak
  - Mn, very weak
  - Others no appreciable stabilization
- To Decrease Eutectoid Content
  - Mn, Cr, V, considerably
  - Ni, Mo, W, somewhat
  - Si, Cu, Al, no appreciable effect

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[1] BULLENS, D. K. and BATTELLE MEMORIAL INSTITUTE, "Steel and Its Heat Treatment," Vol. II, p. 196, John Wiley & Sons, Inc., New York, 1939.

CHAPTER XVII

INDIVIDUAL EFFECTS OF ALLOYING ELEMENTS

IRON-NICKEL ALLOYS

The present important use of nickel as an alloying element in steel dates back to about 1890. Since that time the development and investigation of its use has been very rapid, so that it now finds an exceedingly wide field of industrial application.

Constitution of Iron-Nickel Alloys.—The diagram (Fig. 1-XVII) gives the composite result of the work of several investigators. It shows very definitely that nickel lowers the transformation both on heating and on cooling. In fact, with a nickel content of between 33 and 34 per cent, the Ar_3 point has been lowered to temperatures well below room temperature. In steels of below 34 per cent nickel, the gamma phase will on cooling transform to the alpha phase but is very sluggish in making the transformation, as is indicated by the broad field in which both the alpha and the gamma phase are present. Similar action is noted on heating these alloys. Because of this marked hysteresis, the equilibrium temperature position is not known and is indicated by the temperature ranges that have been observed on slow cooling and heating.

The alloys containing more than 34 per cent nickel exhibit only a magnetic change, as is indicated by the line so marked on the diagram. This transformation is very similar to the A_2 transformation of iron and, therefore, is not a true transformation. These alloys remain as gamma solid solution from the point of solidification to normal temperatures.

It is evident from this discussion that the alloys below 34 per cent nickel can be heat-treated in much the same manner as the iron-carbon alloys.

General Effects of Nickel in Steel.—Nickel apparently dissolves in all proportions in both the gamma and the alpha iron phases. In the presence of carbon, the nickel will remain in solution in the gamma phase of the steel and in the alpha phase at temperatures below the critical range. Since it apparently does not combine with carbon to form nickel carbides but remains in solution in the ferrite, it will strengthen that phase and make an excellent base for the addition of alloys of the carbide-forming group. The principal elements used with nickel are chromium, molybdenum, vanadium, silicon, and manganese. This intensifying action of nickel, which is particularly effective with chromium, produces alloy steels with higher elastic ratios, greater hardness, and higher impact and fatigue resistance than is possible with the simple alloy steels.

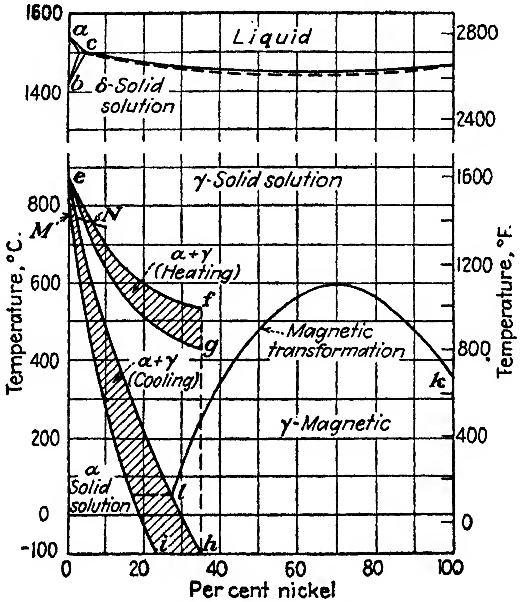

FIG. 1-XVII.—Iron-nickel equilibrium diagram. (*After* 1939 "*Metals Handbook.*")

Nickel, as has been noted, has a very important influence on the strength and ductility of the steel. In the annealed or untreated condition it will increase the strength and elastic limit over a corresponding straight carbon steel, with little decrease in ductility. Nickel steels are much more responsive to heat-treatment, especially hardening, and will in general promote greater toughness. In the heat-treated condition nickel will increase either the strength or the toughness of the steels. Thus, if a 3 per cent nickel steel is heat-treated so that the tensile strength is similar to the strength of a plain carbon steel, the nickel steel will be much tougher. Nickel, when added to both cast and wrought steel, also increases or retains its toughness or impact resistance at low temperatures.

The critical rate of cooling of nickel steels is found to be considerably lowered, which means that, as a result, hardening is more easily accomplished even in thick sections. This property also gives assurance that there will be less danger of warping and distortion in quenching, owing to the fact that less drastic quenching media may be used to produce the desired properties.

High Nickel Alloys.—Nickel, when present in amounts above 20 per cent and in combination with other alloying elements, has many varied and fascinating properties, such as the following:

1. A minimizing of the coefficient of expansion—a property that makes it particularly valuable in precision instruments and dimensional standards.
2. Valuable magnetic characteristics in the very high nickel alloys, which make them applicable in radio work and telephone equipment.
3. Increased resistance to corrosive media, oxidation resistance at high temperatures, and high strength at elevated temperatures. Practically all the steels in these fields are austenitic and not hardenable by heat-treatment.

IRON-CHROMIUM ALLOYS

Chromium as an alloying addition to simple steels, thereby producing a chromium steel, was the first alloy steel to be commercialized. Its development was a distinctively American achievement, as it has been commercially produced since 1869, when Julius Bauer established the Chrome Steel Works in Brooklyn. These early steels were noted for their unusual hardness, and steels of similar type exist in our modern alloys for the same purpose.

Chromium is probably the most universal in its application of all the alloying elements. In addition to the simple chromium steels, which may range in carbon from 0.10 to well over 1.00 per cent, we have chromium in conjunction with nearly all the other elements, *e.g.*, chromium-nickel, chromium-vanadium, chromium-tungsten, chromium-tungsten-vanadium, chromium-molybdenum, and nickel-chromium-molybdenum steels.

Binary System—Fe, Cr.—Before considering the iron-carbon-chromium system let us study the influence of chromium on iron. Figure 2-XVII[1] is the constitutional diagram of the iron end of the binary iron-chromium diagram. As can be seen, chromium in the absence of carbon definitely affects the critical points. As the chromium is increased, the transformation A_4 for pure iron is constantly lowered, while the A_3 is less affected, being at first gradually lowered and then gradually raised until the two points A_3 and A_4 merge to form the so-called *gamma loop*. This merging occurs at about 12 per cent chromium and 88 per cent iron. In other words, the temperature of occurrence of the delta-to-gamma transformation has been lowered and corresponds to the same temperature at which the alpha-to-gamma transformation takes

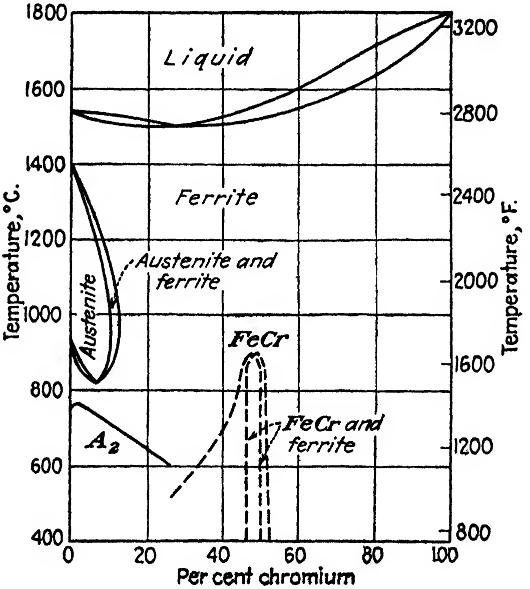

FIG. 2-XVII.—The binary iron-chromium diagram. (*After* 1939 "*Metals Handbook.*")

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[1] KRIVOBOK, V. N., Constitution of Chromium and Chromium-Nickel Steels in "The Book of Stainless Steels," 2d ed., edited by E. E. Thum, Chap. III, pp. 19–49, A.S.M., Cleveland, Ohio, 1935.

place. Accordingly, if we take an alloy of 6 per cent chromium and cool it from a high temperature, the delta chromium iron will transform to chromium austenite at the A_4 critical point, which has been lowered by the chromium addition to around 1225°C. Upon further cooling, the chromium austenite will transform at the A_3 critical point to chromium ferrite, which phase is stable on further cooling to room temperature. There will, of course, be a magnetic change. These alloys are all responsive to heat-treatment.

If a carbon-free alloy of more than 13 per cent chromium is cooled slowly from a high temperature, austenite will no longer be formed. The delta chromium iron, which has formed on solidification, merely transforms to alpha chromium ferrite, which remains stable from that temperature to room temperature. As would be expected, this steel is little affected by heat-treatment; the only change will occur when it has been annealed for a long time at high temperatures—a process that increases the grain size and decreases the toughness.

Three-component System—Fe, Cr, C.—With carbon additions to the iron-chromium system, the situation becomes considerably more complicated. Equilibrium conditions for the iron-carbon-chromium system have not been adequately worked out, but there is definite knowledge at hand regarding the trends that the transformations will follow when chromium is added in varying amounts to the iron-carbon alloys. These trends, which will be discussed, are taken from the works of V. N. Krivobok.

Figure 3-XVII-*B* shows the cross-sectional diagram for iron-carbon alloys containing 6 per cent chromium. In iron-carbon alloys the ordinate represents iron with definite percentages of carbon additions. In this diagram the intersections of the line representing 6 per cent chromium, 0 per cent carbon (Fig. 2-XVII), with the transformation points, have been transferred to the 0 per cent carbon ordinate; the entire ordinate, then, represents carbon additions to this alloy.

Similarly, in Figures 3-XVII-*C* and *D* the 0 per cent carbon ordinate is now 12 per cent chromium, 88 per cent iron, 0 per cent carbon and 18 per cent chromium, 82 per cent iron and 0 per cent carbon, respectively.

The various regions of the iron-chromium-carbon alloys should be readily understood from analogy with the iron-carbon diagram.

From the diagram representing 6 per cent chromium, it is plain that the chromium addition has shifted the eutectoid point *S* to a much lower value for carbon. The point *E* represents the maximum solubility of carbon in gamma chromium-iron and is not definitely placed, but it has been shifted. This shifting changes the position of the A_{cm} to lower carbon regions and results in narrowing the gamma field. The temperature of the A_1 critical transformation is raised slightly and from the upward slope to line *PS* the indications are that it is no longer independent of the carbon content, although the available data indicates that this influence is slight. The position of the A_1 in chromium alloys of relatively high carbon content is somewhat in doubt, owing to the extreme sluggishness of the alloys; in other words, the hysteresis effect on heating and cooling is so pronounced that the establishment of the true A_1 point is extremely difficult. Therefore, beyond the eutectoid point *S* the position of the A_1 line is only tentative. A similar condition exists with regard to line *PL*, which also was drawn without having definite values ascribed to it. The line is important insofar as it shows the relationship between the phases. For purposes of heat-treatment the lines of interest are *GS* and *PSK*. It is interesting to note, in this diagram and succeeding ones shown, that, instead of having pearlite formed, theoretically at least, at a constant temperature, as in iron-carbon alloys, there is an interval of temperatures *PSKL*, where pearlite forms and where three phases—alpha iron, gamma iron, and complex carbides—are present under equilibrium conditions.

The diagram representing the steels containing 12 per cent chromium represent the class in which the alloys become "stainless." The phases and relative positions of the various regions remain the same, but the phase boundaries are shifted to different temperatures and concentrations, and the phase compositions are altered. The A_4 critical has now been lowered to such an extent that it joins the A_3 at about 1000°C. (1830°F.). The A_1 critical in comparison to the 6 per cent chromium alloys has been raised still further. With an increase in carbon up to the eutectoid point *S*, the temperature of occurrence of the A_1 is distinctly raised. Beyond the eutectoid point the same condition of sluggishness, as was previously noted, makes the limits (*PSKL*) difficult to establish. The eutectoid point *S*, in this case, has been further shifted to about 0.35 per cent carbon. The point *E*, which represents the maximum solubility of carbon in gamma chromium iron, is not definitely fixed; but there has been a greater relative shift than in the 6 per cent chromium steels and the

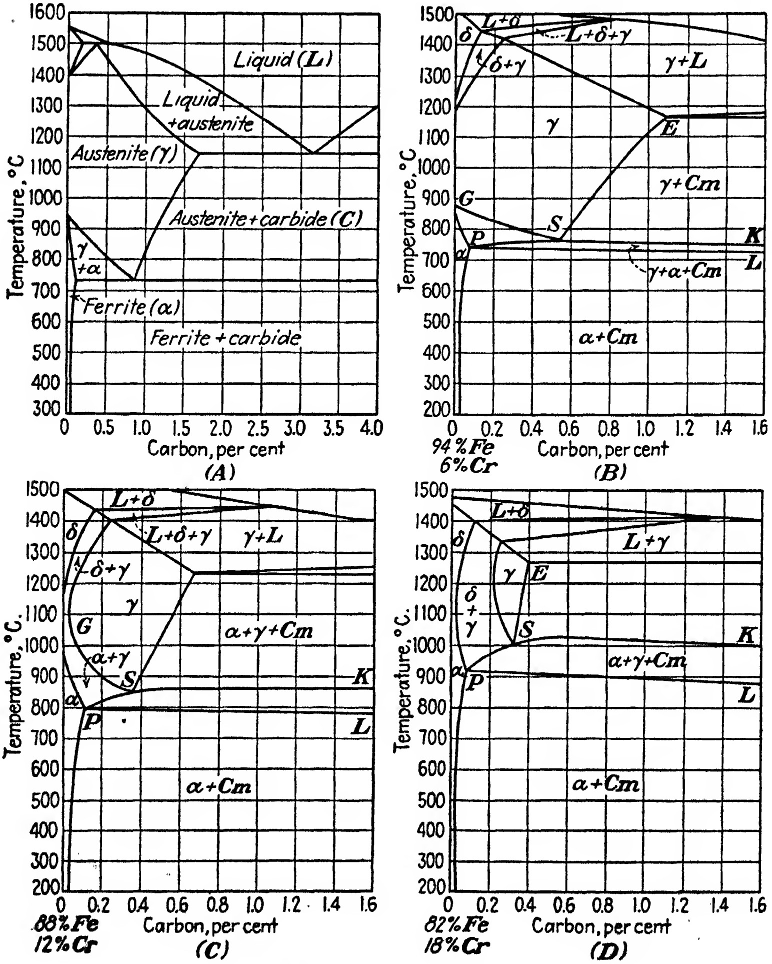

FIG. 3-XVII.—*A*, phase diagram for iron carbon alloys. *B*, cross section of the phase diagram for iron-chromium-carbon alloys taken at 6 per cent chromium. *C*, cross section of the phase diagram for iron-chromium-carbon alloys taken at 12 per cent chromium. *D*, cross section of the phase diagram for iron-chromium-carbon alloys taken at 18 per cent chromium. (*After V. N. Krivobok.*)

point *E* is now not far from 0.70 per cent carbon. The A_{cm} line is, therefore, shifted to lower carbon contents and this results in further narrowing of the gamma field.

Taking an alloy containing about 0.10 per cent carbon let us trace, theoretically at least, the changes that occur on slow cooling. At a temperature of 1000°C., there should be one phase present, *i.e.*, gamma. Upon further cooling, this phase transforms to chromium ferrite and chromium iron carbide. After the formation of the chromium ferrite it can be seen that with decreasing temperature its solubility for carbon decreases. Alloys of greater than 0.35 per cent carbon will, upon cooling, precipitate excess carbides and consist of gamma and carbides to the eutectoid temperature, after which the usual transformation takes place to alpha iron, which has a little chromium and carbide in solution with iron-chromium carbides.

It is evident from these considerations that most of the alloys of fairly low, medium, or high carbon content will be affected by heat-treatment, since equilibrium cannot be established at room temperature in these alloys without a phase change. Thus, if an alloy of this group is quenched, a martensitic structure will result. If it is slowly cooled, the steel will show ferrite and carbides. The very low carbon alloys are little affected by heat-treatment.

The diagram representing those steels containing 18 per cent chromium or those steels outside of the gamma loop shows further modifications over the 12 per cent chromium; the phases, however, remain the same. The most important modification occurs in the region of the eutectoid. The range of temperatures over which the eutectoid transformation takes place has been widened considerably and also raised. Therefore, the rate of transformation has become increasingly slow. The temperature of the transformation depends considerably on the rate of either heating or cooling, but even under very favorable conditions it is seldom completed. The upper limit of carbon concentration (*PS*) is definitely affected by an increase in carbon. For steels with higher carbon content, above eutectoid percentage, the transformation of the austenite becomes increasingly sluggish and makes the limits (*PSKL*) difficult to establish. The eutectoid point has been further shifted to about 0.25 per cent carbon, while the maximum solubility of carbon in gamma iron is now around 0.40 to 0.50 per cent. The gamma field is narrowed still further through this effect.

These alloys have the necessary prerequisites for hardening. The only difference is in the degree of response or the tendency to undercool before the transformation has had an opportunity to take place. It should be noticed that this effect became increasingly predominant as the alloy content was increased.

In the alloys of very low carbon content, corresponding to the narrow field in which chromium ferrite is present, heat-treatment is of little value, which is as might have been predicted. Therefore, these steels cannot be hardened. However, with increasing carbon, the gamma phase makes its appearance and the presence of this phase ensures response to heat-treatment. When the carbon content is gradually increased, the hardenability of the steel increases but never reaches a high figure until the carbon content exceeds 0.25 to 0.30 per cent. This is due to the fact that in the low carbon steels there is always precipitated a certain amount of the alpha phase, which will not harden, owing to the lack of a phase change. In the higher carbon steels, with sufficiently high heating the steel enters fully into the field where the gamma plus cementite is present; this ensures response to heat-treatment.

General Effects of Chromium in Steel.—Chromium is found to dissolve and be retained in solid solution in both the gamma and the alpha phase of iron. Chromium, however, is primarily a carbide former and, when sufficient carbon is present, the chromium combines with it to form carbides or double carbides of iron and chromium. In the low carbon steels or after the carbon has been satisfied, it goes into solution in the ferrite.

The eutectoid carbon content is found to be lowered by chromium additions, the amount varying with the per cent of chromium present. The position of the Ac_3 varies according to the chromium and carbon contents. When large amounts of chromium are present the Ac_3 is materially raised, but in low percentages when sufficient carbon is present it may be lowered.

The carbides of chromium are apparently difficult to take into solution in austenite. Thus, unless there is allowed sufficient time at the heating temperature, the steels are likely to remain fine-grained, owing to the low rate of solution of the carbides.

Chromium steels, on the whole, will require higher heating temperatures for heat-treatment. When there has been sufficient solution of chromium in austenite, the critical cooling rate is decreased and hence the hardening power and the depth of penetration are increased.

As the chromium content of the low carbon steels is increased beyond about 10 per cent, the resistance to chemical attack by specific media is very greatly increased. This resistance is probably due to the formation of a very thin, stable film on the surface of the metal, which serves as a barrier to chemical attack by specific media. This resistance is especially true with respect to oxidation resistance at both low and high temperatures. The oxidation at elevated temperatures increases with increased chromium content beyond about 10 per cent and is very rapid as the chromium exceeds 20 per cent. In addition, nickel and other alloying elements increase the resistance to attack by media that are reducing rather than oxidizing.

One of the outstanding effects of the chromium-bearing steels is their wear resistance and cutting ability. This particular characteristic is due to the hardness of the chromium-bearing carbides.

Chromium is found also to contribute somewhat to the strength at high temperatures. However, in such applications it is ordinarily used in conjunction with other alloys, such as molybdenum and tungsten.

The physical properties of chromium steels also appear to be quite sensitive to the balance between the chromium and carbon, particularly the properties of the high-chromium steels. If too high a carbon content is used, there will be little ferrite strengthening and the steels will tend toward brittleness. In general, the chromium steels for a given strength will have a higher yield ratio and a somewhat lower ductility than some of the other alloys. Since chromium is essentially a hardening element rather than a toughener, it is often used in conjunction with a toughening element to produce the desirable characteristics.

The chemical properties also are greatly affected by the carbon content. In general, the higher the chromium content and the lower the carbon, the more resistant the alloy will be toward certain types of corrosion. Any change in this relationship will tend to lower the corrosion resistance.

IRON-CHROMIUM-NICKEL ALLOYS

Constant investigation with modified chromium alloys has shown that nickel additions are the most important. The use of nickel in combination with high chromium has resulted in alloys that are resistant not only to atmospheric corrosion but also to oxidation at high temperatures. To render these alloys resistant to high-temperature oxidation, the nickel content is required to be around 8 per cent with the chromium around 18 per cent. This is the outstanding commercial steel of the so-called *Strauss* type known as 18-8. The carbon content of these steels is generally quite low, ranging between 0.05 and 0.20 per cent maximum. Other combinations of chromium and nickel will also yield good results, and many different compositions are now manufactured. The largest tonnage of this grade of steel is made with a carbon content of 0.10 per cent and below.

Constitutional Diagram.—Now that we are familiar with the effect of nickel on iron and of chromium on iron, and the effect of carbon on each, let us investigate the complex system—chromium, nickel, iron, carbon, and temperature. It is apparent that this system is complex and impossible to represent as a space diagram. Most of the investigation in this field of alloys has been carried on with alloys of approximately 18 per cent chromium to which varying amounts of nickel have been added and carbon up to 0.50 per cent.

Starting with an 18 per cent chromium, 0 per cent carbon alloy, let us study the effect of adding varying amounts of nickel. The structural diagram is shown in Fig. 4-XVII. The iron-18 per cent chromium alloy, as is known from previous study, will consist of one phase, chromium-alpha solid solution. With increasing additions of nickel a second phase makes its appearance, *i.e.*, gamma solid solution or chromium-nickel-austenite. This austenite, it can be seen, is stable only at elevated temperatures and will on cooling undergo either a partial transformation, the alloys thus consisting of alpha plus gamma solid solution, or complete transformation to alpha alone, depending upon the composition. If carbon is present in these alloys, even in very small amounts, the equilibrium conditions are considerably complicated by the presence of carbides, either as usually found or in solution.

For a comparison of the chromium-nickel alloys with the chromium alloys, we will use sections that correspond to give constant chromium and nickel content, with carbon and temperature as the variables.

18 Chromium-4 Nickel Steels.—Figure 5-XVII-*A* shows the cross-sectional diagram of the 18-4 steels with varying amounts of carbon. In making the comparison it should be noticed that the horizontal scale has been greatly magnified. It is at once evident that the nickel acts on the iron-chromium alloys in the same way that it does on iron. Nickel lowers the transformation temperatures and increases the field in which the gamma phase is present. Thus, it can be seen that nickel, when present in sufficient quantities, exerts the same influence on the iron-chromium alloys as it does on pure iron. It can also be seen that in these alloys the reaction austenite → ferrite is extremely sluggish (as is indicated by the dotted area where the three phases are present) and hence there is a pronounced tendency for undercooling before the phase change has had an opportunity to

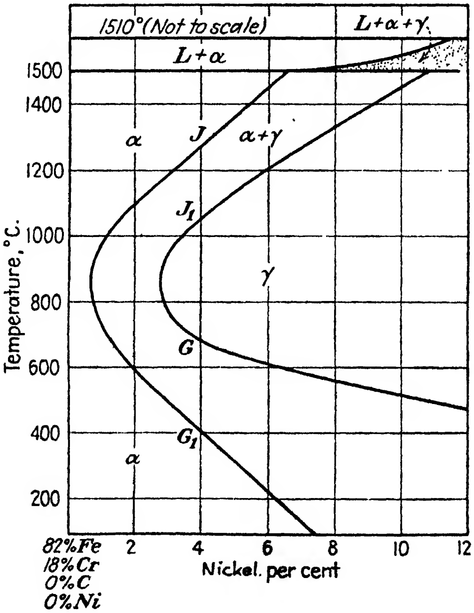

FIG. 4-XVII.—Structural diagram for carbon free alloys of 82 per cent iron, 18 per cent chromium, variable nickel. (*After V. N. Krivobok.*)

occur. An alloy of sufficiently high carbon—say, 0.30 per cent—may be retained in the metastable gamma state at room temperature by means of fairly rapid cooling. On the above metastable austenite's being heated somewhat above room temperature, it will gradually transform; but the rate of transformation will be slow.

It must be noted that, although we are considering this diagram as a more or less distorted iron-carbon diagram in regard to concentrations and temperatures of various transformations, it does not follow that we shall find an exact similarity in the microscopic appearance of these alloys as compared with plain carbon steels. For example, an alloy cooled through the three-phase region SK will not show the typical pearlite of carbon steels but will show the phases distributed at random, with the carbides finely dispersed throughout. This formation is due to the complicated eutectoid transformation, which occurs over a range of temperatures and causes a constant readjustment of phase compositions, and to the sluggishness, or delayed mode of transformation.

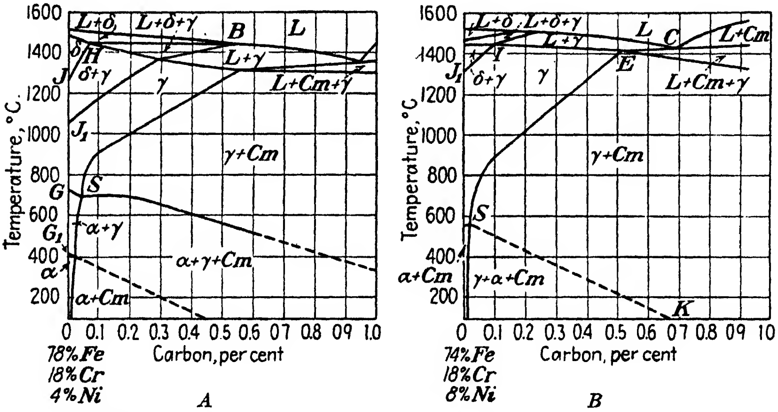

FIG. 5-XVII.—*A*, influence of carbon on iron alloyed with 18 per cent chromium and 4 per cent nickel. *B*, trend of the reactions in steels alloyed with 18 per cent chromium and 8 per cent nickel. (*After V. N. Krivobok.*)

The constituents of these alloys depend entirely upon heat-treatment and this will be reflected in their physical properties. With sufficient carbon present, they can be made either fairly hard or relatively soft. The very low carbon alloys cannot be heat-treated and, hence, will have no appreciable hardness.

18 Chromium-8 Nickel Steels.—Figure 5-XVII-*B* shows the cross-sectional diagram of the 18-8 steels with varying amounts of carbon. The increased addition of nickel has now, by comparison, further shifted the phase boundaries to different temperatures and concentrations and, as before, owing to the sluggishness, the position of the phase lines is quite indefinite.

In the industrial alloys of this type it is found that by quenching from a temperature of from 1000 to 1200°C. (1830 to 2190°F.) they will remain austenitic, with all the carbide in solution. In this condition they are the softest and most ductile. Cold work is found to be one of the most effective ways of accelerating the delayed transformation; after cold work the alpha phase, as well as the carbide, will appear.

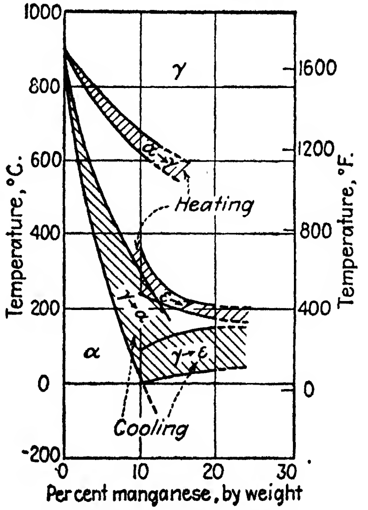

FIG. 6-XVII.—Iron end of the iron-manganese equilibrium diagram. (*After* 1939 "*Metals Handbook*.")

IRON-MANGANESE ALLOYS

The use of manganese as an alloying element, particularly its use in small quantities, has developed rapidly following Guillet's experiments, which completely disproved the belief that low manganese steels were brittle. They are now being used extensively. The higher manganese-carbon steels, as originally developed by Robert Hadfield in 1882, have been used extensively since that time and are given an important application in places where there is considerable heavy wear.

Groups of Iron-Manganese Alloys.—The transformation and phases involved in this system are shown in the tentative diagram (Fig. 6-XVII) constructed in accordance with dilatometric evidence. The diagram shows only the iron end of the iron-manganese diagram, as this end is of most importance to our present discussion. On this end of the diagram we can see that manganese acts very much like nickel, in that the critical trans-

formations are materially lowered; and it causes a very pro nounced hysteresis effect between the heating and cooling transformations, i.e., the alpha to gamma on heating and the gamma to alpha on cooling.

The alloys of the first group to be noted are those that vary in composition from 0 to 10 per cent of manganese. On cooling, alloys in this group will undergo the gamma-to-alpha transformation, which is complete at room temperatures. These alloys contain only one phase at room temperature, *i.e.*, alpha iron.

The second group consists of those alloys in which the gamma phase transforms to epsilon, with or without the formation of alpha. This epsilon phase has only recently been discovered and as yet there is insufficient evidence concerning its nature. Its presence is shown by dilatometric and X-ray studies and can be observed microscopically when the manganese becomes sufficiently high to suppress the gamma-to-alpha transformation (about 16 per cent).

The third group is composed of those alloys containing from 30 to 60 per cent of manganese, in which the transformations have all been completely suppressed and the gamma phase is present from high temperatures to room temperature and below.

It is very apparent that the alloys of the first group can be heat-treated in a way similar to that for iron-carbon alloys. The second group shows the existence of a Widmanstätten type of structure under all conditions of treatment. The third group, the gamma-to-alpha transformation, has been completely suppressed, so that they are not changed by heat-treatment and, therefore, remain austenitic.

Iron-Carbon-Manganese Alloys.—The phase diagrams[1] shown in *A*,*B*,*C*, and *D*, of Fig. 7-XVII represent cross sections of the ternary iron-carbon-manganese system for alloys containing of 0.00, 2.50, 7.00 and 13.0 per cent manganese. These diagrams show the phases that are in equilibrium at various temperatures as the carbon is increased in the iron-manganese alloy designated. They do not, however, tell the concentrations of manganese and carbon or the relative amounts of phases in any of the heterogeneous fields.

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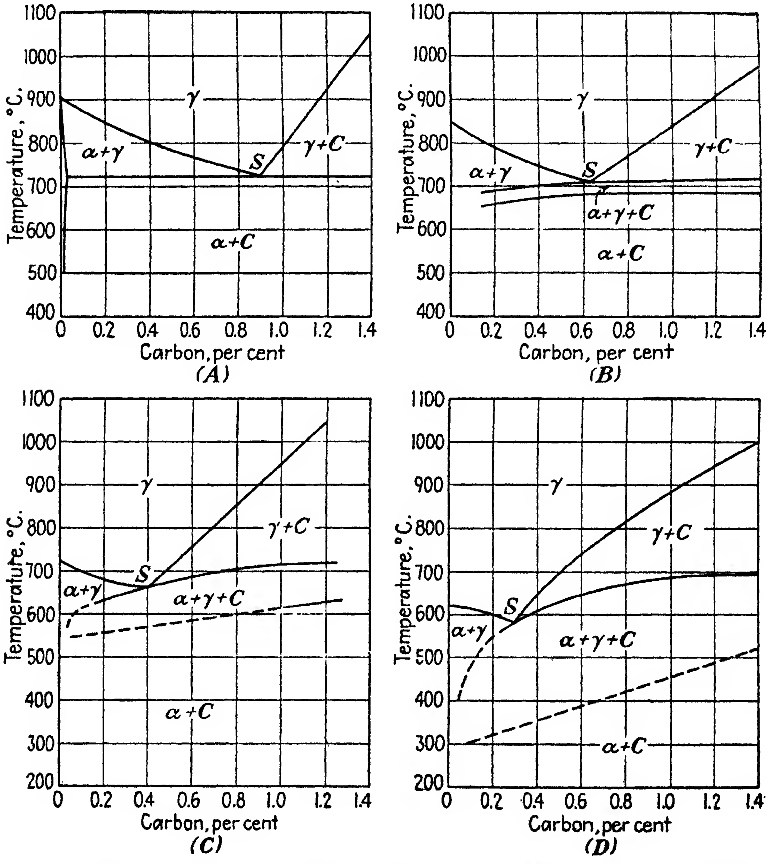
[1] WELLS, C., Constitution of Iron-manganese-carbon Alloys, A.S.M. "Metals Handbook," pp. 409–415, Cleveland, Ohio, 1939.

In general, it can be seen that manganese addition to steel lowers the transformation temperatures and tends to render such transformations extremely sluggish. That is, the allotropic transformations do not take place at any definite temperature but occupy a whole range of temperatures. This follows the general principle of alloy steels, which do not possess a single-temperature critical point on heating for both the disappearance



(A) (B)

(C) (D)

FIG. 7-XVII.—*A*, phase diagram for iron-carbon alloys. *B*, cross section of the phase diagram for iron-manganese-carbon alloys taken at 2.50 per cent manganese. *C*, cross section of the phase diagram for iron-manganese-carbon alloys taken at 7 per cent manganese. *D*, cross section of the phase diagram for iron-manganese-carbon alloys taken at 13 per cent manganese. (*After Walters and Wells, Trans. A.S.M.*)

of ferrite or carbide (hypo- or hypereutectoid steels, respectively) and the formation of austenite. The A_1 designation for such steels, therefore, carries a degree of inaccuracy depending upon the amount and nature of the alloying elements. It further can be seen that the eutectoid temperature is lowered and that the point is shifted to the left as the manganese content is increased.

It must again be noted that the estimation of the composition of the phases present in any part of these diagrams cannot be made. From a practical point of view this is not important, as the useful information concerns the temperature and composition limits of the various fields of the diagrams.

General Effects of Manganese in Steel.—Manganese is present in practically all of our commercial steels and is essential to them during their manufacture, acting indirectly in the capacity of a scavenger. It reacts with the sulphur to form manganese sulfide, which is then slagged off. It affects the degasifying and deoxidizing of the molten steel during the melting, teeming, and solidifying stages and thus contributes to the production of sound steel. As a result of this manganese addition, plus a safety-factor percentage, in steels there is always a remanent manganese content, which runs as a minimum around 0.30 to 0.40 per cent. It is difficult to designate the exact percentage of manganese above which the quantities may be considered definite alloy additions. It would seem that when the manganese is present in quantities around 1 per cent or greater the steels are regarded as manganese steels, with the exception of those alloys to which manganese is added as an alloying element for a specific purpose, such as 18-8 plus manganese in amount up to 2 per cent, etc.

Manganese when added to steel is found to be moderately soluble in alpha iron, even more soluble in gamma iron, and forms manganese carbide, Mn_3C, which is found associated with the cementite. By virtue of these facts, the ultimate strength is increased and, within certain limits, its toughness is also increased.

The eutectoid percentage of carbon is lowered to about 0.78 per cent for 1 per cent manganese and to about 0.67 for 2 per cent manganese. The effect of very high manganese on the eutectoid is shown in the diagrams for iron-carbon-manganese alloys in Fig. 7-XVII. In the 1 per cent range, manganese is found to lower the Ac_3 about 115°, with no further drop up to 3 per cent.

Manganese very markedly decreases the critical cooling rate and, therefore, contributes to deep hardening. With reference to the S-curve, this means that manganese not only lowers the position of the curve but also shifts it to the right.

Several of the difficulties involved in such steels are: (1) The use of too high carbon and too high manganese produces brittleness, and (2) the steels have a tendency to have a coarse grain size. The first difficulty can be eliminated by using a proper balance between the carbon and manganese contents, while the latter can be controlled by the addition of grain-controlling elements such as vanadium or titanium.

With very high percentages of manganese with high carbon, the steels are particularly resistant to impact types of abrasion.

IRON-VANADIUM ALLOYS

Vanadium did not enter the metallurgical field until about 1896, at which time it was used experimentally in the manufacture of armor plate at the Firminy Steel Works, France. It was then definitely observed that the vanadium additions produced superior plates, but this element was not given a great deal of attention as an alloying element until the investigations by Arnold in 1900–1901. In 1904 Sankey and Smith presented their investigations on chromium-vanadium steels, and this really marks the beginning of vanadium steel as a commercial product.

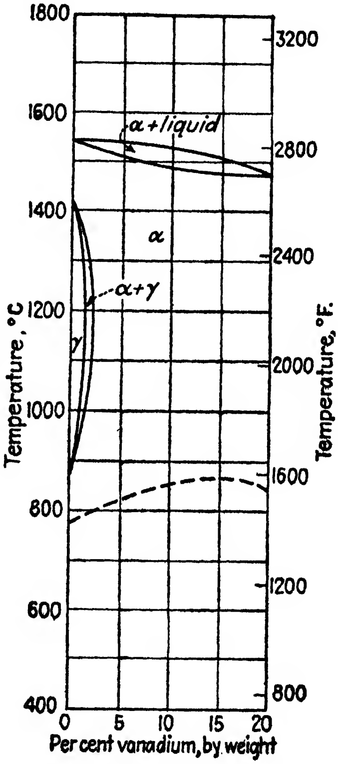


FIG. 8-XVII.—Iron end of the iron-vanadium equilibrium diagram. (*After* 1939 "*Metals Handbook.*")

Constitutional Diagram.—The approximate iron-vanadium constitutional diagram is shown in Fig. 8-XVII. The iron-rich end of the diagram is shown, as this is of most interest to our present discussion. As can be seen, vanadium lowers the A_4 critical and raises the A_3 critical and results in a merging of the two points to form the gamma loop. A heterogeneous field also exists in which the gamma and alpha phases coexist in equilibrium with each other. The boundaries of the gamma and the gamma and alpha fields are not as yet definitely fixed; but, according to the works of Wever and Jellinghaus, the limit of the gamma field is placed at 1.00 per cent of vanadium with the limit of the gamma and alpha field at 1.10 per cent. Since these alloys have the necessary transformation points, they can be heat-treated.

With increasing amounts of vanadium above 1.10 per cent, there exists merely a solid solution of vanadium in alpha or delta iron.

General Effects of Vanadium in Steel.—Vanadium may be used to influence the properties of steel indirectly through its use as a deoxidizing agent and directly as an addition to steel. It is seldom used as a deoxidizing agent, because of its high cost; to use it for this purpose would be poor economy when other cheaper and very effective elements are available. It is, however, added to the steel for its alloying qualities after the deoxidation has been effected by use of the cheaper alloys.

Vanadium is found to dissolve to some degree in the ferrite, where it imparts added strength and toughness, but it is essentially a carbide-forming element, in which capacity it increases strength and hardness. The carbides thus formed are quite stable and show little tendency to segregate or form large masses, thus stabilizing the structures in which they are present. This stability is further evident at high temperatures.

Vanadium steels in the "as-forged" or "as-rolled," annealed, normalized, or quenched-and-tempered condition will show a much finer structure than similarly treated steels of a similar composition without vanadium. The grain-growth tendencies are also minimized at temperatures in the heat-treating range, thus allowing much higher hardening or normalizing temperatures. This factor also markedly widens the quenching ranges and produces a series of steels that will withstand abusive handling in heat-treating.

Vanadium is the most effective alloy in lowering the eutectoid ratio. For every 1 per cent vanadium added, the eutectoid ratio is lowered about 0.21 per cent.

Vanadium not only improves the tensile strength with no loss in ductility but it also increases the elastic limit, yield point, and impact strength.

It also has the effect of intensifying the individual properties of the other major elements in the more complex steels.

IRON-MOLYBDENUM ALLOYS

Molybdenum, as an addition to steel, is perhaps the latest member of the group of alloying elements. Investigations of this element as an alloy were reported for the most part in the latter part of the nineteenth century. The works of Guillet in 1905 and Swinden in 1911 and 1913 were most important in calling attention to the beneficial effects of molybdenum in constructional steels. Later works by Mills, who recognized and patented the advantages obtained by the use of molybdenum in constructional steels, and the studies by Guillet and Mack in 1925, are noteworthy. The commercial use of steels containing molybdenum, particularly chromium-molybdenum steels, has advanced rapidly. The actual recognition of the commercial importance of molybdenum came about in 1926, when it was placed in the S.A.E. listing as the 4100 series.

Constitutional Diagram.—The approximate iron-molybdenum constitutional diagram is shown in Fig. 9-XVII. Only the iron-rich end of the diagram is shown, as this is of most interest to our present discussion. As can be seen, molybdenum addition to iron lowers the A_4 critical and raises the A_3 critical and, as a result, they merge to form the gamma loop. Sykes[1] in constructing his diagram estimated the limit of the gamma field to be 3 per cent molybdenum. Therefore, all alloys falling within this limit have the alpha-to-gamma and gamma-to-alpha transformation on heating and cooling and can be heat-treated.

With increasing additions of molybdenum, as is shown in the diagram, the solid solution of molybdenum in alpha or delta iron coexists from the solidus to room temperature. There is a region, however, in which the alpha solid solution exists in equilibrium with the intermetallic compound Fe_3Mo_2. The solubility of molybdenum at different temperatures is noted by the phase boundary between the ferrite field in which ferrite

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[1] SYKES, W. P., Constitution of Iron-Molybdenum Alloys, A.S.M. "Metals Handbook," pp. 384–385, Cleveland, Ohio, 1939.

coexists with Fe_3Mo_2. The temperature of formation of the intermetallic compound and the maximum solubility of molybdenum is shown on the diagram.

General Effects of Molybdenum in Steel.—It is found that molybdenum either can exist in solution in the iron matrix or can combine with the carbon to form complex carbides. Its tendency to form carbides is even greater than that of chromium. Owing to these properties, molybdenum steels will have high strength, high elastic limit, resistance to wear, and high impact qualities.

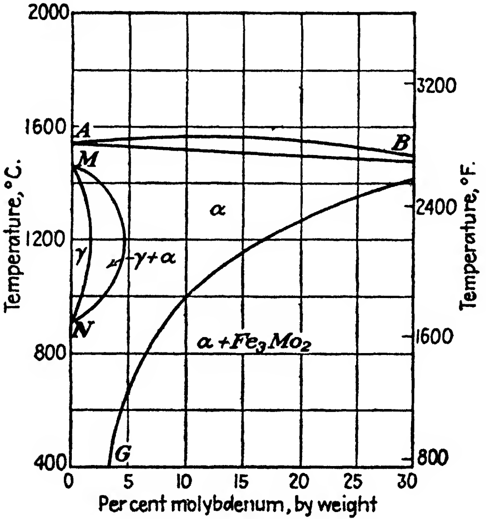

FIG. 9-XVII.—Iron end of the iron-molybdenum equilibrium diagram. (*After* 1939 "*Metals Handbook.*")

The eutectoid composition is found to decrease with increasing amounts of molybdenum, so that at about 1 per cent the eutectoid is approximately 0.80 per cent. When added in the usual amounts for constructional purposes (0.20 to 0.30 per cent) the Ac_3 is slightly raised.

The carbides of molybdenum are similar to those of vanadium and chromium, in that they are quite stable and hence will assist in restraining grain growth. As a result of the slow solubility and the rise in the Ac_3, the temperatures for hardening are usually higher than they would be for steels without the molybdenum.

When some of the molybdenum is in solution in the austenite, the reaction rates for the transformation of austenite become markedly decreased; hence, these steels are quite deep-hardening.

It has been found that molybdenum steels in the as-quenched condition require a much higher tempering temperature to be softened to the same degree as other comparable alloy steels. This resistance to tempering indicates another property—that of showing ability to retain their strength at elevated temperatures. For this same reason they show resistance to "creep" under sustained loads below their elastic strength at elevated temperatures.

Molybdenum also contributes to the red-hardness property, or ability to withstand heating to a visible heat without a material lowering of the hardness. Combined with chromium, nickel, and vanadium, it finds wide application in hot-working tools and dies. The action of molybdenum in steels is generally accepted as being similar in character to that of tungsten, but the effect of molybdenum differs in magnitude, 1 per cent of molybdenum being equivalent to 2 to 3 per cent of tungsten.

It should be noted that molybdenum shows its best influences when used in combination with other alloys.

IRON-TUNGSTEN ALLOYS

Alloy steels, including those containing tungsten, were produced more or less experimentally in the early part of the nineteenth century. The first commercially important alloy steel was the "self-hardening" steel containing tungsten and manganese developed by Mushet in 1868. Later developments in the latter half of the nineteenth century on self-hardening steels were toward reducing the manganese and adding chromium. In 1898 Messrs. Taylor and White started investigating Mushet and other self-hardening steels and found that by proper heat-treatment at high temperatures these steels could be made many times superior to steels as they were usually hardened. Taylor and White are generally credited with the discovery of high-speed steels, but their discovery was really the method of treatment and not the steel itself. As soon as the results of their investigations became known, considerable investigation and experimenting was started, which resulted in the discovery that the old self-hardening steels could be appreciably improved by changing their composition. This has resulted in today's high-speed steels, which differ from the old compositions in that they contain less carbon and more tungsten.

Tungsten magnet steels were extensively produced and used from the latter part of the nineteenth century to the time of the First World War. Since then they have been largely replaced by the cheaper chromium steels or by the more effective cobalt steels.

In addition to its use in high-speed steels, tungsten is used in many tool and die steels and in heat-resisting steels. It is generally used in combination with other alloying elements.

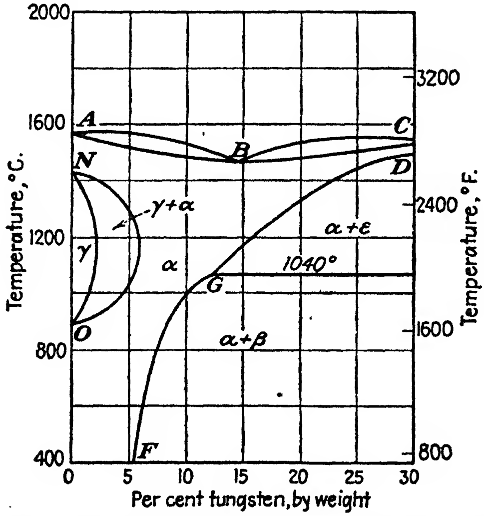

FIG. 10-XVII.—Iron end of the iron-tungsten equilibrium diagram. (*After* 1939 "*Metals Handbook.*")

Constitutional Diagram.—The approximate constitutional diagram of the binary iron-tungsten system is shown in Fig. 10-XVII. The diagram shows only the iron end of the iron-tungsten diagram, as this end is of most importance to our present discussion. Tungsten as can be seen acts very much like chromium, molybdenum, and vanadium, in that the A_4 critical is lowered by increasing amounts of tungsten, while the A_3 critical points are raised. As a result there is an intersection with the formation of the gamma loop. As can be seen in the diagram, a heterogeneous field exists, where gamma and alpha or delta phases are present and are in equilibrium with each other. The maximum solid solubility of tungsten is approximately 6 per cent at 1200°C. Alloys containing less than 3 per cent tungsten are completely transformed to the gamma phase on being heated or cooled through this region. Alloys containing 3 to 6 per cent are only partly transformed into the gamma phase. Since there is such a transformation, these alloys have the necessary requisites for heat-treating and can, therefore, be recrystallized, have grain refinement, and have their hardness increased.

With increasing amounts of tungsten, above 6 per cent, and depending upon the temperature, the following phases will be present: (1) α solid solution of tungsten in delta or alpha iron; (2) α solid solution plus the intermetallic compound $\epsilon(Fe_3W_2)$; and (3) α solid solution plus the intermetallic compound $\beta(Fe_2W)$. As would be expected, the iron-rich solid solutions containing more than 6 per cent tungsten (solid solubility at 500°C.) are susceptible to precipitation hardening.

General Effects of Tungsten in Steel.—In behavior tungsten is found to be similar to molybdenum. It is a gamma loop-forming element, raises the Ac_3 critical transformation, and somewhat decreases the eutectoid carbon percentage.

The complex carbides formed by tungsten are found to be stable and, as a result, the quenching temperatures are high and the time for solution is extremely long. This stability tends to retard grain growth and, as a result, widens the allowable quenching range. If tungsten is taken into solution in the austenite, it tends to decrease the critical cooling rate and will, therefore, have strong depth-hardening properties.

Owing to the fact that tungsten goes into solution in the iron matrix and also forms carbides, it will tend to increase the strength, hardness, and toughness of the steel.

The stability of the carbides is further evidenced on tempering the as-quenched steels; they require much higher temperatures to produce a given softness than do carbon steels. This reluctance toward tempering also bears a relation to the retention of strength at high temperatures. However, since molybdenum is stronger in its effect on this property and has a lower cost, and since a much smaller quantity is needed to produce a comparable result, it is universally used in preference to tungsten where load-carrying ability at high temperatures is needed.

The particular properties of heat resistance and wear resistance of the tungsten steels are utilized in the fast-finishing class of tool steels, which utilize high carbon, with around 3 to 6 per cent tungsten. These same properties are utilized in the complex combination of tungsten and other elements, such as chromium, vanadium, and molybdenum. In the very high tungsten steels, which contain around 14 per cent tungsten, the austenite and martensite formed in the hardened steels are so extremely reluctant to transform on tempering that only slight softening occurs up to 700°F. (owing to the martensite decomposition), while from 700 to about 1100°F. the hardness increases (owing, apparently, to the transformation of the remaining austenite to martensite). This particular characteristic of secondary hardness is termed *red hardness.*

IRON-SILICON ALLOYS

From the diagram now proposed for the iron-silicon alloys[1] it is found that silicon raises the A_3 and lowers the A_4 to form a gamma loop, with a limit of 2.2 per cent silicon. When additions of carbon are made, the per cent of silicon necessary to make the alloys ferritic increases; as, for example, with 0.10 per cent carbon and over about 3 per cent silicon, the alloys are ferritic; with 0.30 per cent carbon and over 7.5 per cent silicon, the alloys are ferritic.

The critical temperature on heating, Ac_2, is raised with increasing silicon content; the amount varying with the carbon content. The heating temperature necessary for the usual treatments of normalizing and quenching will, therefore, be increased.

Since silicon is not a carbide-forming element but enters into solution in the ferrite, it will contribute to an increase in the ultimate strength without sacrificing ductility or toughness when added in amounts up to 2.50 per cent. With amounts above this figure, silicon produces brittleness and, with amounts above 5 per cent, renders the steel or iron nonmalleable. As a ferrite strengthener, silicon is used in some of the low alloy high-yield strength steels.

In some of the complex steels, silicon is added for the improvement it develops in oxidation resistance at high temperatures. It does not have an effect on the high-temperature strength.

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[1] STOUGHTON and GREINER, The Constitution of Iron-silicon Alloys, A.S.M. "Metals Handbook," p. 396, Cleveland, Ohio, 1939.

TABLE 1-XVII.—SPECIFIC EFFECTS OF ALLOYS IN STEEL*

Element	Influence exerted through carbides			Influence on ferrite	Solubility		Principal functions
	Carbide forming tendency	Effect on hardenability	Action during tempering		In gamma iron	In alpha iron	
Aluminum Al	Less than Fe (Graphitizes)	Mildly positive (If any dissolved in austenite)	....	Hardens considerably by solid solution	1.1 % (Increased by C)	30 % ±	1. Deoxidizes efficiently 2. Restricts grain growth (by forming dispersed oxides or nitrides) 3. Alloying element in nitriding steel
Chromium Cr	Greater than Mn Less than W	Moderate (Up to 1 per cent more effective than Mn)	Mildly resists softening	Hardens slightly (increases corrosion resistance)	12.8 % (20 % with 0.5 % C)	Unlimited	1. Increases corrosion and oxidation resistance 2. Increases hardenability 3. Contributes some strength at high temperatures 4. Resists abrasion and wear (with high carbon)
Cobalt Co	Similar to Fe	Negative (That is, decreases it)	Sustains hardness by solid solution	Hardens by solid solution	Unlimited	80 % ±	1. Contributes to red hardness by hardening ferrite
Manganese Mn	Greater than Fe Less than Cr	Moderate ($\mathrm{Ni} < \mathrm{Mn} < \mathrm{Cr}$) Retains austenite	Very little, in unusual percentages	Hardens markedly and reduces plasticity	Unlimited	15 to 18 %	1. Counteracts embrittlement by inevitable sulfur 2. Increases hardenability inexpensively 3. Forms batter resistant steel (high Mn, high C)
Molybdenum Mo	Strong Greater than Cr	Great (When dissolved) $\mathrm{Mo} > \mathrm{Cr}$	Opposes softening by a secondary hardening	Gives age-hardening systems in high Mo-Fe alloys	3 % ± (8 % with 0.30 % C)	32 % (Less with lower temperatures)	1. Raises coarsening temperature of austenite 2. Deepens hardening 3. Raises hot strength, creep resistance, and red hardness 4. Enhances corrosion resistance in stainless
Nickel Ni	Less than Fe (Graphitizes)	Mild Retains austenite in medium carbon steels	Very little in small percentages	Strengthens and toughens by solid solution	Unlimited	25 % ± (Irrespective of carbon content)	1. Strengthens unquenched or annealed steels 2. Toughens pearlitic-ferritic steels (especially at low temperature) 3. Renders high chromium-iron alloys austenitic

* Adapted from E. C. Bain's "The Function of Alloying Elements in Steel," A.S.M., Cleveland, Ohio, 1939.

TABLE 1-XVII.—SPECIFIC EFFECTS OF ALLOYS IN STEEL.*—(*Continued*)

Element	Influence exerted through carbides			Influence on ferrite	Solubility		Principal functions
	Carbide forming tendency	Effect on hardenability	Action during tempering		In gamma iron	In alpha iron	
Phosphorus P	Nil	....	....	Effective hardener by solid-solution effect	0.5 %	2.5 % (Irrespective of carbon content)	1. Strengthens low carbon steel 2. Increases resistance to corrosion 3. Improves machinability in free-cutting steels
Silicon Si	Negative (Graphitizes)	Moderate	Sustains hardness by solid solution	Hardens with loss in plasticity Mn < si < P	2 % ± (9 % with 0.35 % C)	18.5 % (Even with considerable carbon)	1. Used as general purpose deoxidizer 2. Alloy for electrical and magnetic sheet 3. Improves oxidation resistance 4. Increases hardenability of steels carrying nongraphitizing elements 5. Strengthens low alloy steels
Titanium Ti	Maximum (2 per cent Ti renders 0.50 per cent carbon steel unhardenable)	Strong, if dissolved (Persistent carbides restrain grain growth and withhold carbon; therefore reduce hardenability)	Persistent carbides probably unaffected	Age-hardening systems in high Ti-Fe alloys	0.75 % (1 % with 0.20 % C)	6 % ± (Less with lower temperatures)	1. Fixes carbon in inert particles *a.* Reduces martensitic hardness and hardenability in medium chromium steels *b.* Prevents formation of austenite in high chromium steels *c.* Prevents localized depletion of chromium in stainless steels during long heating
Tungsten W	Strong	Considerable, especially in small percentages	Opposes softening by secondary hardening	Age-hardening systems in high W-Fe alloys	6 % ± (11 % with 0.25 % C)	32 % (Less with lower temperatures)	1. Forms hard, abrasion-resistant particles in tool steels 2. Promotes red hardness and hot strength
Vanadium V	Very strong V < Ti or Cb	Marked (when dissolved)	Maximum for secondary hardening	Hardens moderately by solid solution	1 to 2 % (4 % with 0.20 % C)	Unlimited	1. Elevates coarsening temperature of austenite (promotes fine grain) 2. Increases hardenability (when dissolved) 3. Resists tempering 4. Gives marked secondary hardening

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* Adapted from E. C. Bain's "The Function of Alloying Elements in Steel," A.S.M., Cleveland, Ohio, 1939.

Probably one of the most important effects of silicon in steels lies in its use as a deoxidizer and degasifier in molten steels and thus contributes toward the production of sound steel. In this capacity it is usually used up to about 0.20 per cent in the killed wrought steels and up to 0.40 per cent in castings.

One of the most important uses of straight silicon steels comes about through its specific effect of developing electrical and magnetic properties, such as high magnetic permeability and low hysteresis loss, which are desirable in steels used for transformers, dynamos, etc. There are a number of grades of silicon steels for this purpose, ranging from about 0.25 per cent to 4 per cent or more of silicon, all with very low carbon contents.

Another of silicon's specific effects is its ability to make cementite unstable and thus form graphite. In cast irons the production of either gray iron, in which the graphite is precipitated as flakes, or white iron, in which the carbon is all combined as cementite, depends upon the proper balance of carbon and silicon; high carbon, high silicon produces gray iron, while a reduction in these amounts produces white iron. In the high carbon steels, silicon is used in amounts ranging around 1 per cent to produce strong, easily machinable steels that can be forged or rolled instead of having to be cast, as in cast iron and malleable iron, for application where wear resistance is important. The fine-tempered carbon produced by proper heat-treatment induces extremely good machinability and makes the alloy similar to cast iron, in freedom from galling and for service conditions where there is rubbing with little lubrication. This latter steel is one of a group of the so-called *graphitic* steels.

SUMMARY OF THE SPECIFIC EFFECTS OF ALLOYS IN STEEL

Table 1-XVII will serve as a summary of the specific effects of the various alloys in steel with respect to their solubility in gamma and alpha iron, their influence exerted through carbides, their influence on ferrite, and their principal functions.

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CHAPTER XVIII

S.A.E. ALLOY STEELS—CARBURIZING GRADES

In our previous study of the plain carbon carburizing grades, it was found that they were low carbon steels that were applied, after carburizing and proper heat-treatment, in services that demanded a hard, wear-resisting surface and a tough shock-resisting core. The alloy carburizing grades are also low carbon steels, which must serve in the same capacity but more efficiently than the simple grades. For different applications, the user has a number of different grades of steel and a number of different carburizing operations from which to choose. The choice of the type of carburizing steel will be determined by the application or service requirements, balanced against the cost and treatment of the steel.

The following S.A.E. carburizing steels will be discussed in reference to their properties, applications, heat-treatment, and properties developed following heat-treatment: S.A.E. 2315, 2515, 3115, 3312, 4615, 4815, and 6115.

It is found that for light parts or in cases where an extremely tough core is required, the carbon content is preferably 0.18 per cent maximum, while for heavy parts requiring strong cores the carbon content should be from 0.15 to 0.25 per cent.

In regard to carburizing these steels, we know from the previous studies that the temperature and time selected for carburization will depend upon the type of carburizer, the steel, the characteristics desired in the case, the equipment available, etc.—all of which must be regulated to suit the conditions.

The rate of penetration of carbon in carburizing is somewhat less in the nickel and nickel-molybdenum steels than in the nickel-chromium steel. Service requirements will, however, determine the final depth of the hardened case. Proper allowance as usual should be made for machining and for warpage. The alloy content is not the only variable that affects the penetration. Grain size is found to be a factor, as well—*i.e.*, coarse-grained

TABLE 1-XVIII*

S.A.E. number	Carbon range	Manganese range	Phosphorus (max.)	Sulfur (max.)	Nickel range	Chromium range	Molybdenum range	Vanadium range	On slow heating		On slow cooling	
									Ac_1	Ac_3	Ar_3	Ar_1
2315	0.10–0.20	0.30–0.60	0.04	0.05	3.25–3.75	....	....	....	1300	1440	1350	1100
2515	0.10–0.20	0.30–0.60	0.04	0.05	4.75–5.25	....	....	....	1250	1420	1220	825
3115	0.10–0.20	0.30–0.60	0.04	0.05	1.00–1.50	0.45–0.75	....	....	1355	1500	1470	1240
3312	Max. 0.17	0.30–0.60	0.04	0.05	3.25–3.75	1.25–1.75	....	....	1330	1435	1240	1160
4615	0.10–0.20	0.40–0.70	0.04	0.05	1.65–2.00	....	0.20–0.30	....	1335	1485	1400	1200
4815	0.10–0.20	0.40–0.60	0.04	0.05	3.25–3.75	....	0.20–0.30	....	1330	1440	1310	800
6115	0.10–0.20	0.30–0.60	0.04	0.05	....	0.80–1.10	....	0.15	1420	1550	1450	1300

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* S.A.E. "Handbook," 1940; MORRIS, SERGESON, and GABLE, Approximate Critical Temperatures for S.A.E. Steels, *Metal Progress*, **28** (No. 2), 40-A (1935).

steels, according to the McQuaid-Ehn test, are found to carburize a little more rapidly than fine-grained steels.

Grain size is another factor when the proper heat-treatment is under consideration. The typical double-quench method brings about grain refinement by using a high temperature to refine the core and to dissolve the free carbides in the case, and a second lower temperature to refine the high carbon case and at the same time temper the low carbon core.

With the advent of fine-grained steels, one quench has been found to give satisfactory results. This single quench is made from the higher temperature used in the double quench. Thus it is possible to obtain refinement of the core with a minimum coarsening of the case and minimum distortion by a single quench. The three popular grain-controlled alloy, carburizing steels are the S.A.E. 3115, S.A.E. 4615, and S.A.E. 4815.

In regard to machining of fine-grained steels it is found that they do not always machine as readily as the coarse-grained steels without being specially normalized. It seems that the users of the fine-grained steels are now asking for steels that resist grain growth at the carburizing temperatures (1600 to 1700°F.) so that they will be tough in the as-treated condition and show a McQuaid-Ehn grain size of 5 to 7, but will coarsen rapidly at higher normalizing temperatures to provide good machining qualities.

Since the core is relatively low in carbon, there will be very little internal stress set up by the quenching, thus making it unnecessary to use a high-temperature draw to secure a tough, finished part. The case, however, will be stressed; hence, it is recommended that a low-temperature draw (not over about 350°F.) be used to relieve some of the stresses and still maintain wear hardness.

Analysis and Critical Temperatures.—The chemical composition and approximate critical temperatures for the typical carburizing grades of alloy steels are shown in Table 1-XVIII. It should be noted that the silicon content of all the S.A.E. basic open-hearth alloy steels shall be 0.15 to 0.30 per cent. In the electric and acid open-hearth alloys, the silicon shall be 0.15 per cent minimum.

Physical Properties.—The physical properties of the typical carburizing grades of alloy steels are shown, in Table 2-XVIII, both in the "as-rolled" condition and following the typical single-quench treatment employed when grain-controlled steels are used. For further reference in regard to heat-treatment and physical properties for these grades of steel the following handbooks are suggested: S.A.E. "Handbook," 1940; A.S.M. "Metals Handbook," 1939; "Agathon Alloy Steels," Republic Steel Corporation; "Bethlehem Alloy Steels," Bethlehem Steel Company, 1935.

TABLE 2-XVIII.*—PHYSICAL PROPERTIES (CORE)† OF CARBURIZING STEELS

Steel	As rolled		Following heat-treatment‡		
	Tensile p.s.i.	Brinell	Tensile p.s.i.	Brinell	Izod ft.-lb.
S.A.E. 2315	80,000–100,000	156–207	181,000	375	38–40
S.A.E. 2515	85,000–120,000	187–241	190,000	387	50–53
S.A.E. 3115	80,000–100,000	156–207	156,000	340	34–36
S.A.E. 3312	80,000–120,000	156–207	200,000	400	
S.A.E. 4615	80,000–100,000	156–207	134,000	302	45–46
S.A.E. 4815	80,000–115,000	187–241	182,000	375	50–51
S.A.E. 6115	85,000–105,000	179–217	145,000	332	59–64
S.A.E. (6120) Ni-Cr-V	75,000– 95,000	156–196	138,000	302	60–64

* From "Agathon Alloy Steels Handbook," Republic Steel Corporation, and "Bethlehem Alloy Steels Handbook," Bethlehem Steel Company.

† Size treated, 1 in. diameter.
Size tested, 0.505 in. by 2 in.

‡ Heat-treatment, assuming fine-grained steels.
1. Carburized 1650 to 1700°F.—8 hr.
2. Direct oil quench.
3. Temper approximately 300°F.

GENERAL CHARACTERISTICS AND USES

S.A.E. 2315.—This is one of the oldest of the nickel carburizing steels. Nickel as an alloy enters into solution with the ferrite and not only strengthens it, but also improves the ductility of the heat-treated part, has a moderate effect on the hardenability, and aids in grain refinement. The critical points *Ac* and *Ar* are both lowered, thereby allowing lower quenching temperatures and lessening distortion.

This steel is used primarily in kingpins, steering worms, transmission gears, truck ring gears, chain pins, etc. Figure

1-XVIII shows the typical structures of this steel in the as-rolled condition, and the case and core after carburizing followed by a quench from 1650°F.

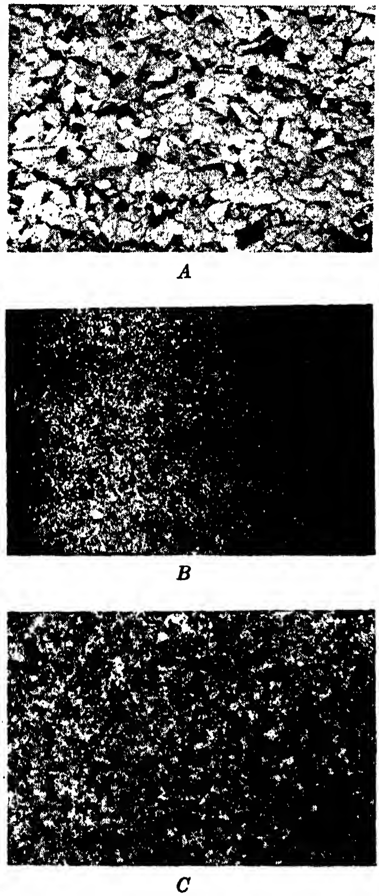

Fig. 1-XVIII.—Typical structure S.A.E. 2315 steel. *A*, as rolled × 100. *B*, carburized, single quench from 1650°F. Case × 100. *C*, core × 100.

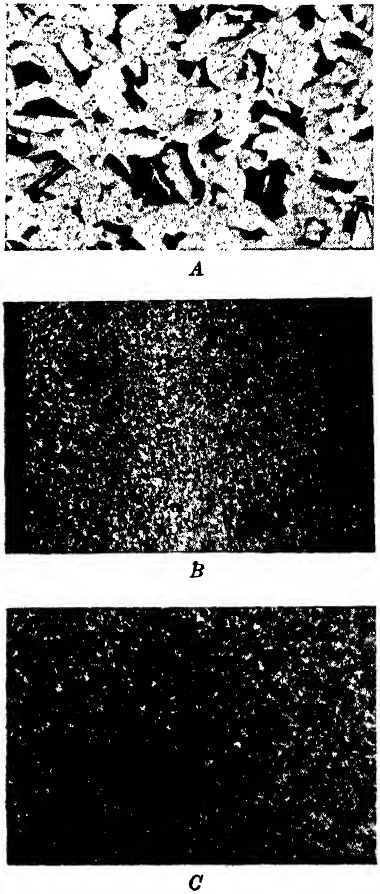

Fig. 2-XVIII.—Typical structure S.A.E. 6115 steel. *A*, as rolled × 100. *B*, carburized, single quench from 1650°F. Case × 100. *C*, core × 100.

(*Courtesy of Republic Steel Corporation.*)

S.A.E. 2515.—In this higher nickel carburizing steel, the ability of nickel to develop toughness and high physical properties is very well shown. This steel gives an excellent, hard-wearing surface, backed by an extremely strong, ductile, and shock-resistant core after heat-treatment. The surface hardness will probably not be as great as that of some of the other steels because of the retained austenite that is often present in the hardened case. With this high nickel content, the critical temperatures are so lowered that a minimum amount of distortion will be ensured, thus making this a desirable steel for high-grade carburizing work.

This steel is especially applicable for heavy service, such as truck or bus gears, spline shafts, piston pins, countershafts, drive gears, airplane-engine parts, etc.

S.A.E. 3115 (3120).—This steel is one in which have been incorporated recent improvements in the physical chemistry of steelmaking, to produce a grain-controlled steel. The combination of nickel and chromium has produced a steel in which the chromium increases the hardening capacity while the nickel influences the strength and ductility.

This steel finds its use, after carburizing, in such parts as rock bit cones, pins, worms, gears and pinions of many types, universal joints, rolls, rocker arms, airplane-engine cams, studs, bolts, etc.

S.A.E. 3312.—This steel, with higher nickel content, is called *Krupp analysis*, and represents one of the finest carburizing steels. It has all the advantages imparted by the combination of nickel and chromium. As would be suspected, it has very high strength and toughness and is unusually excellent for applications that require resistance to high pressure, shocks, and abrasion. With suitable heat-treatment, these steels provide very high core strengths and, owing to their air-hardening properties, are especially suited for large sections intended for heavy-duty service.

This steel finds its application in truck and bus transmission gears, large roller bearings and races, airplane engines, piston pins and gears, etc.

S.A.E. 4615 (4620).—Molybdenum as an alloying element in steel is most commonly used to enhance the properties of other alloying elements and bring about a much-improved response to heat-treatment. This particular grade, in which there is a combination of nickel and molybdenum, shows, after carburizing, high fatigue resistance and high tensile properties.

The fine-grained types of this steel show very tough cores, with a single quench in oil. This fact has proved a very valuable asset in the production of parts that require a deep and exceptionally hard case.

This steel has had almost universal use in steel for cups, cones, and rollers in all case-hardened roller bearings except those of large size. In large sections, where distortion and warping are problems and high surface hardness is necessary, this steel has been very satisfactory. It also is given wide use in gears of all types in the automotive field, for journals, spline shafts, wrist pins, pinions, freewheeling spiders, case-hardened mining drill parts, etc.

S.A.E. 4815 (4820).—This steel, which is a fairly recent development, produces a much higher core strength, owing to the effect of the higher nickel content, than does the S.A.E. 4615. These steels are often used interchangeably with the S.A.E. 2515 steels, because they have very closely approaching physical properties.

These steels have proved their value in such applications as gears, freewheeling shells, roller bearings, chain pins, high-strength studs and bolts, oil-well cutter bits, pinions, etc.

S.A.E. 6115 (6120).—Vanadium, as an alloying element, acts as a scavenger for the impurities and has the characteristic property of producing a fine-grained structure. Vanadium, in effect, promotes ductility and accentuates the properties of the other alloying elements. The S.A.E. 6115 (6120) steels have, owing to their inherent fine-grained characteristics, a very desirable application in shops where close pyrometric control is not available.

This steel finds its application, as carburized, in pneumatic-tool parts, wrenches, roller-bearing cones, pistons, and such parts, where the stresses are high and where a high surface hardness is required.

Figure 2-XVIII, shows the typical structures of this steel in the as-rolled condition and the case and core after carburizing, following a quench from 1650°F.

CHAPTER XIX

S.A.E. ALLOY STEELS—WATER- AND OIL-HARDENING STEELS

GENERAL PROPERTIES AND TREATMENT

The oil- and water-hardening steels differ from the carburizing steels in that they do not depend upon a high carbon surface for their properties. In general, the steels with a carbon content of less than about 0.30 to 0.35 per cent are water-hardening steels on parts that have section sizes between ¾ and 2¼ in., while those of higher carbon content are the oil-hardening steels.

Analysis and Critical Temperatures.—The chemical composition and the approximate critical temperatures for the typical water- and oil-hardening steels are shown in Table 1-XIX. The silicon content, unless otherwise noted, for the S.A.E. basic open-hearth grades will be 0.15 to 0.30 per cent, while in the electric and acid open-hearth alloys, the silicon shall be 0.15 per cent minimum.

Heat-treatment and Physical Properties.—The typical annealing, normalizing, and quenching temperatures for the water- and oil-hardening steels are shown in the first part of Tables 2-XIX and 3-XIX. These tables also show the approximate physical properties for the water- and oil-hardening steels in the annealed and in the quenched-and-drawn at 800°F. conditions.

For further reference in regard to heat-treatment and physical properties for these grades of steel the following references are suggested: S.A.E. "Handbook," 1940; A.S.M. "Metals Handbook," 1939; "Agathon Alloy Steels Handbook," Republic Steel Corporation; "Bethlehem Alloy Steels Handbook," Bethlehem Steel Company, 1935.

MEDIUM MANGANESE

Manganese, when used as an alloying element in steel, improves the strength in the "as-rolled" condition, and, when heat-treated, improves the combination of strength and ductility, provided that the section is not too large. In the heavier sections, the higher alloy steels will respond more satisfactorily.

TABLE 1-XIX*

S.A.E. number	Carbon range	Manganese range	Phosphorus (max.)	Sulfur (max.)	Nickel range	Chromium range	Vanadium	Silicon range	Molybdenum range	On slow heating		On slow cooling	
										Ac_1	Ac_3	Ar_3	Ar_1
T1330	0.25–0.35	1.60–1.90	0.04	0.05	....	....	....	....	....	1325	1480	1340	1160
1340	0.35–0.45	1.60–1.90	0.04	0.05	....	....	....	....	....	1315	1435	1310	1160
2330	0.25–0.35	0.50–0.80	0.04	0.05	3.25–3.75	....	....	....	....	1275	1400	1180	1050
2345	0.40–0.50	0.60–0.90	0.04	0.05	3.25–3.75	....	....	....	....	1280	1350	1180	1060
3130	0.25–0.35	0.50–0.80	0.04	0.05	1.00–1.50	0.45–0.75	....	....	....	1345	1460	1360	1220
3145	0.40–0.50	0.60–0.90	0.04	0.05	1.00–1.50	0.45–0.75	....	....	....	1355	1395	1295	1220
3240	0.35–0.45	0.30–0.60	0.04	0.05	1.50–2.00	0.90–1.25	....	....	....	1335	1425	1280	1240
4130	0.25–0.35	0.50–0.80	0.04	0.05	....	0.50–0.80	....	....	0.15–0.25	1395	1485	1403	1280
4145	0.40–0.50	0.60–0.90	0.04	0.05	....	0.80–1.10	....	....	0.15–0.25	1385	1470	....	1275
4340	0.35–0.45	0.50–0.80	0.04	0.05	1.50–2.00	0.50–0.80	....	....	0.30–0.40	1350	1425	875	725
4645	0.40–0.50	0.50–0.80	0.04	0.05	1.50–2.00	....	....	....	0.20–0.30	1320	1400	....	1200
5130	0.25–0.35	0.60–0.90	0.04	0.05	....	0.80–1.10	....	....	....	1370	1470	....	1290
5145	0.40–0.50	0.60–0.90	0.04	0.05	....	0.80–1.10	....	....	....	1365	1450	....	1275
6130	0.25–0.35	0.60–0.90	0.04	0.05	....	0.80–1.10	0.15	....	....	1390	1485	1370	1285
6145	0.40–0.50	0.60–0.90	0.04	0.05	....	0.80–1.10	0.15	....	....	1390	1450	1375	1290
9255	0.50–0.60	0.60–0.90	0.04	0.05	....	....	....	1.80–2.20	....	1400	1500	1380	1320

* S.A.E. "Handbook," 1940; MORRIS, SERGESON, and GABLE, Approximate Critical Temperatures for S.A.E. Steels, *Metal Progress*, **28** (No. 2), p. 40-A (1935).

TABLE 2-XIX.*—WATER-HARDENING STEELS

Steel S.A.E. number	Temperature, °F.			Physical properties,† anneal		Quench temperature as shown—draw at 800°F.				
	Anneal F.C.	Normalize A.C. (typical)	Quench, water	Tensile, p.s.i.	Brinell	Tensile, p.s.i.	Yield, p.s.i.	Reduction of area, per cent	Elongation, in 2 in. per cent	Brinell
T1330	1450	1650–1750	1525	90,000–100,000	192	174,000	154,000	43	16	340
2330	1475	1625–1725	1475	80,000– 90,000	149–196	170,000	150,000	53	17	355
3130	1450–1500	1650–1750	1500	80,000– 90,000	149–196	181,000	156,000	52	15	365
4130	1450–1500	1575–1650	1600	80,000– 90,000	156–207	173,000	152,000	55	18	360
5130	1550	1600	1525	75,000– 85,000	143–192	188,000	170,000	49	13	390
6130	1550	1600	1550	80,000– 95,000	156–207	192,000	175,000	48	14	440

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* From "Agathon Alloy Steels Handbook," Republic Steel Corporation, and "Bethlehem Alloy Steels Handbook," Bethlehem Steel Company.

† Size treated, 1-in. round. Size tested, 0.505 in. by 2 in.

TABLE 3-XIX.*—OIL-HARDENING STEELS

Steel S.A.E. number	Temperature, °F.			Physical properties,† anneal		Quench temperature as shown—draw at 800°F.				
	Anneal F.C.	Normalize A.C. (typical)	Quench, oil	Tensile, p.s.i.	Brinell	Tensile, p.s.i.	Yield, p.s.i.	Reduction of area, per cent	Elongation, in 2 in., per cent	Brinell
T1340	1450	1600–1700	1500	100,000–110,000	197	176,000	149,000	42	11	352
2345	1475	1550	1450	80,000–115,000	163–217	200,000	163,000	53	15	388
3145	1525	1600	1500	80,000–115,000	163–217	192,000	175,000	47	13	400
3240	1500	1575	1500	100,000–110,000	197	212,000	190,000	47	14	425
4145	1525	1600	1550	85,000–115,000	170–228	212,000	190,000	47	13	428
4340	1500	1575	1525	90,000–135,000	187–269	225,000	206,000	48	13	450
4645	1500	1575	1475	80,000–120,000	179–235	200,000	181,000	49	15	420
5145	1525	1600	1525	85,000– 95,000	156–207	206,000	187,000	42	12	415
6145	1550	1625	1575	85,000–115,000	179–217	210,000	190,000	42	12	430
9255‡	1550	1600	1575	110,000–120,000	229	225,000	200,000	33	10	460

* From "Agathon Alloy Steels Handbook," Republic Steel Corporation, and "Bethlehem Alloy Steels Handbook," Bethlehem Steel Company.

† Size treated, 1-in. round. Size tested, 0.505 in. by 2 in.

‡ Treated in spring section.

Incidentally, the steels classed as commercial manganese steels are those that have a mean range of over 1.00 per cent manganese; below this content they are considered carbon steels.

S.A.E. X 1330.—This steel is typical of the group in which the carbon content is from 0.25 to 0.35 per cent and may, therefore, be classed as a water-hardening steel. It finds its application in general machine parts in both the treated and the untreated condition, because it will produce higher physical properties than the carbon steels of the same carbon content.

S.A.E. X 1340.—This steel is in the group containing over 0.35 per cent carbon and should be oil-quenched rather than water-quenched, because of the danger of cracking, as a result of the relatively high manganese and increased carbon content.

In the 0.30 to 0.40 per cent carbon range, it is put into service, in the heat-treated condition, in such parts as main bearing bolts, engine bolts, connecting rod bolts, studs, timing chain links, brake drums, etc., in the automotive industry, and in general industry it is applied to pressure cylinders, lagging and road machinery, gears, shafting, axles, miscellaneous bolts and studs, and many other machine parts.

NICKEL STEELS

The addition of nickel to steel provides a very dependable steel, which has greatly increased elastic properties and strength as compared with carbon steels, with no loss in ductility or resistance to impact. The action of nickel in lowering critical temperature ranges is very advantageous and makes possible lower quenching temperatures, with a resultant decrease in scaling and distortion. These steels also have lower critical quenching rates and will, therefore, be deeper hardening than carbon steels. The toughening effect of nickel is particularly noticeable in the medium carbon grades.

S.A.E. 2330.—This steel is used in services that require strength and great toughness, as in highly stressed bolts, nuts, turnbuckles, etc., in heat-treated tubing where resistance to torsion and fatigue are important, as in automobile front-axle tubes, propeller-shaft tubes, rear-axle housings, and other structural parts; and in cold-drawn stock where its desirable combination of high tensile strength and ductility, along with good machining properties, bright surface, and accurate dimensions, makes it applicable for important bolts, nuts, studs, keys, etc. With the addition of manganese, ranging from 0.70 to 1.00 per cent, it finds wide application in railroad service in forgings for side rods, main rods, piston rods, axles, etc.

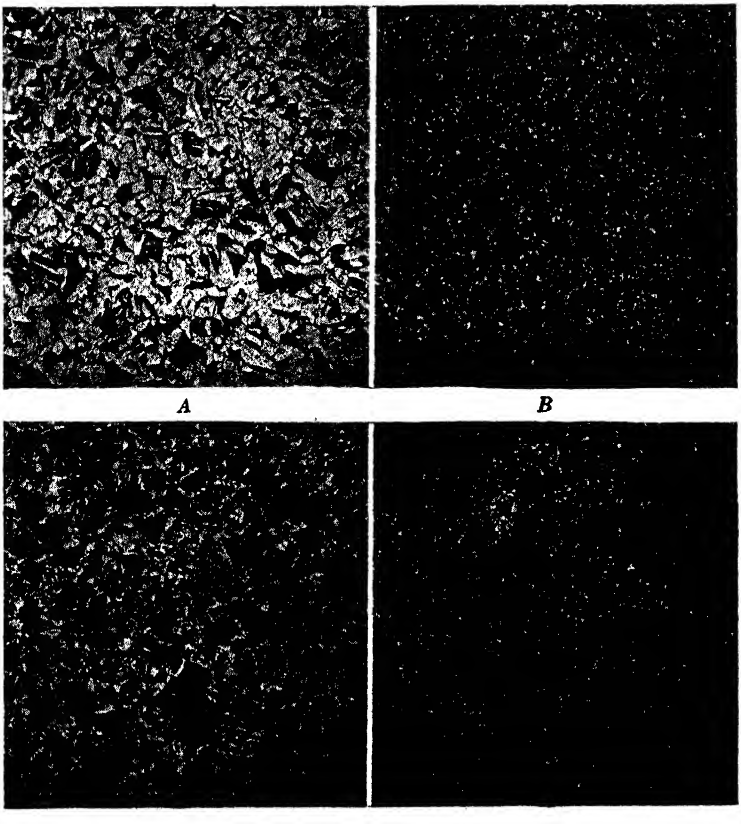

Fig. 1-XIX.—*A*, typical structure of S.A.E. 2330 steel (0.30 carbon, 3.57 nickel) in the annealed condition. × 100. *B*, typical structure of S.A.E. 2330 steel after water hardening and drawing at 800°F. × 100. *C*, typical structure of S.A.E. 2345 steel (0.45 carbon, 3.54 nickel) in the annealed condition. × 100. *D*, typical structure of S.A.E. 2345 steel after oil hardening and drawing at 800°F. × 100. (*Courtesy of Carpenter Steel Company.*)

Figure 1-XIX-*A* and *B* shows the typical structure of this steel in the annealed condition and after water hardening and drawing at 800°F.

S.A.E. 2345.—In the higher carbon range this steel is used in the oil-quenched and tempered condition for oil-hardened axles, drive shafts, steering knuckles and arms, oil-hardened gears, heavy-duty shafting, and highly stressed bolts, units, and machine parts. It is also used, as quenched and drawn, in jar bodies and drill-collar pins in the oil industry.

Figure 1-XIX-*C* and *D* shows the typical structure of this steel in the annealed condition and after oil hardening and drawing at 800°F.

NICKEL-CHROMIUM STEELS

The desirable qualities of the combination of these two elements, in which the chromium increases the hardness penetration and the nickel improves ductility and the strength, are brought out in the following descriptions.

S.A.E. 3130.—This steel is in the water-hardening range but, in complicated shapes, full water quenching is not advisable. By proper study of sections and treatment this steel may be made to produce excellent impact-resisting qualities. This is due to the fact that, in sections up to about 2 in., the strength required permits tempering between 800 and 1200°F. subsequent to hardening, and thus imparts high impact and fatigue resistance with high elastic properties.

Under the conditions just discussed, which give it the ability to withstand alternating stresses up to high limits, this steel is very applicable for axles, drive shafts, highly stressed bolts, nuts, and all parts that require a high factor of safety in design. The S.A.E. 3130 and 3140 steels show very good high-temperature properties. They have a creep resistance, at temperatures up to 1000°F., greatly superior to carbon steels of like carbon content. They are also extensively used in such parts as valves, flanges, and fittings where service temperatures range from 400 to 750°F.

S.A.E. 3145.—This steel is finding very wide use in the oil fields, where an oil-hardening steel is required for rotary drilling equipment. It is also being employed in aircraft, bus and truck crankshafts, power-shovel parts, oil-well tool joints, bit and reamer bodies, etc. When tempered after hardening to about 400 Brinell, it shows excellent wear resistance in road-building, excavating, and farm machinery parts, in contact with nonmetallic abrasive material. It has the advantage over similar carbon grades of having greater toughness and resistance to shock.

S.A.E. 3240.—The S.A.E. 3200 series is most often used in the higher carbon ranges, around 0.40 and 0.50 per cent in cases where somewhat better properties are desired than could be obtained for the S.A.E. 3100 series of similar carbon contents. These steels are very dependable and have as their outstanding characteristic their reliability in heat-treatment. They show a combination of high strength and high ductility (toughness) and, as a result, have application in automobile drive and axle shafts, transmission gears, master connecting rods of radial aircraft engines, oil-well drilling, general tools, gears, etc.

MOLYBDENUM STEELS

The most general use of molybdenum at the present time is to enhance the desirable properties of the other elements, such as manganese, nickel, or chromium, and nickel with chromium, all of which will show, with relatively small additions of molybdenum, an improved response to heat-treatment.

S.A.E. 4130.—This steel was developed to produce a water-hardening steel that would be deep hardening, would not crack during the drastic quench, would produce uniform results when drawn, and would retain high physical properties at high draw temperatures. The chromium itself will produce the above desired characteristics but, with the molybdenum addition, these characteristics are accentuated and the tendency toward temper brittleness is eliminated. One of the additional characteristics of molybdenum steels is their good machining properties after heat-treatment. They can be machined in quantity production with Brinell values as high as 364. The principal uses are for axles, steering knuckles, and steering-knuckle arms; also, for sheet and strip for aircraft work.

S.A.E. 4145.—This steel again presents the combination and balance of chromium and molybdenum that give a steel with excellent hardenability. It is used in parts that require high hardness, especially when the parts are to be machined following heat-treatment, such as axle shafts, wrenches, tools, gears, etc.

S.A.E. 4340.—This steel is an oil-hardening steel with the combination of chromium, nickel, and molybdenum that imparts excellent hardenability, ductility, and impact resistance. It is ordinarily used for axle shafts, but may also be used for other highly stressed parts.

S.A.E. 4645.—This nickel-molybdenum steel, an oil-hardening type, has an alloy content the same as the S.A.E. 4640, but contains slightly higher carbon. It possesses excellent ductility and impact qualities and when heated to low temperatures does not lose its impact resistance as rapidly as do some of the other alloy combinations. It has given excellent service in both leaf and coil springs in railroad service. In the slightly lower carbon content (0.35 to 0.45 per cent), it is used for gears, shafts, and other parts that require high fatigue resistance and physical properties.

CHROMIUM STEELS

Chromium as an addition to steel is essentially a hardener and does not improve ductility. The chromium steels are the S.A.E. 5XXX series, of which the grades S.A.E. 5140 and 5150 are all oil-quenching, deep-hardening types. They find extensive use for springs, gears, shear blades, studs, bolts, agricultural-implement parts, side bars, rollers, rotary and roller chains in the oil fields, etc.

There are many modifications other than those listed in the S.A.E. series that have definite and extensive industrial use. A number of these types containing 0.25 to 0.60 per cent chromium, with carbon contents ranging from 0.45 to 1.20 per cent, are all oil-hardening steels and find extensive use in magnets, axes, mine bits, grinding balls, springs, and wood-cutting tools.

One of the most recent combinations is a steel of carbon about 0.10 per cent maximum, 0.50 to 1.50 per cent chromium, 0.30 to 0.50 per cent copper, 0.10 to 0.20 per cent phosphorus, and 0.50 to 1.10 per cent silicon. The chromium imparts strength and corrosion resistance; the silicon increases the strength without lowering the ductility; the copper aids only slightly in the strength but does impart corrosion resistance; and the phosphorus, in combination with the copper, accentuates the corrosion resistance. Phosphorus in these high percentages usually produces "cold shortness" but, when it is in combination, this undesirable property is not evidenced. This combination produces high strength, good ductility, impact and endurance values, with remarkable resistance to atmospheric corrosion. It has found considerable application in sheet and strip form in such equipment as freight and passenger cars, streamlined trains, streetcars and buses, tank cars, and mine cars.

Another low chromium steel is the so-called *cromansil*, which contains 0.50 per cent chromium, 1.25 per cent manganese, and 0.75 per cent silicon. This steel has been promoted for its excellent strength to weight ratio—a property due to the fact that the silicon raises the yield point, the manganese raises the tensile strength, and the chromium intensifies these effects without altering the excellent ductility and toughness. It has found use in welded engine beds, underframes and other parts used in the construction of high-speed passenger trains, lightweight railroad equipment, dredging buckets, and in boilers and pressure vessels under A.S.M.E. Boiler Code Requirements.

A higher chromium steel containing chromium from 1.85 to 2.70 per cent, in varying ranges, approximately 4.00 per cent silicon and around 0.40 per cent carbon, has found use for intake valves in internal-combustion engines. The high silicon, which has the effect of rapidly raising the *Ar* critical point, retards the air-hardening properties of the steel and thus makes its addition quite valuable in applications where the operating temperature often exceeds the normal hardening or quenching temperature range.

CHROMIUM-VANADIUM STEELS

Vanadium as an alloying addition to steel promotes a finer grain structure, promotes ductility, and accentuates the properties of the other alloying elements, such as manganese and chromium. Its action in widening the forging and heat-treating range makes it very adaptable under conditions of high production, with assurance of uniform results. In combination with chromium, it tends to modify the hardening action and thus improve its toughness or shock-resisting properties.

S.A.E. 6130.—This lower carbon chromium-vanadium steel is of the water-hardening type and has been used very successfully for front axles, steering knuckles, steering arms, crankshafts, connecting rods, drive shafts, axle shafts, steam-hammer piston rods, small gears, etc. The steels containing above 0.30 per cent carbon should not be water quenched.

Figure 2-XIX-*A* and *B* shows the typical structure of this steel in the annealed condition and after water hardening and drawing at 800°F.

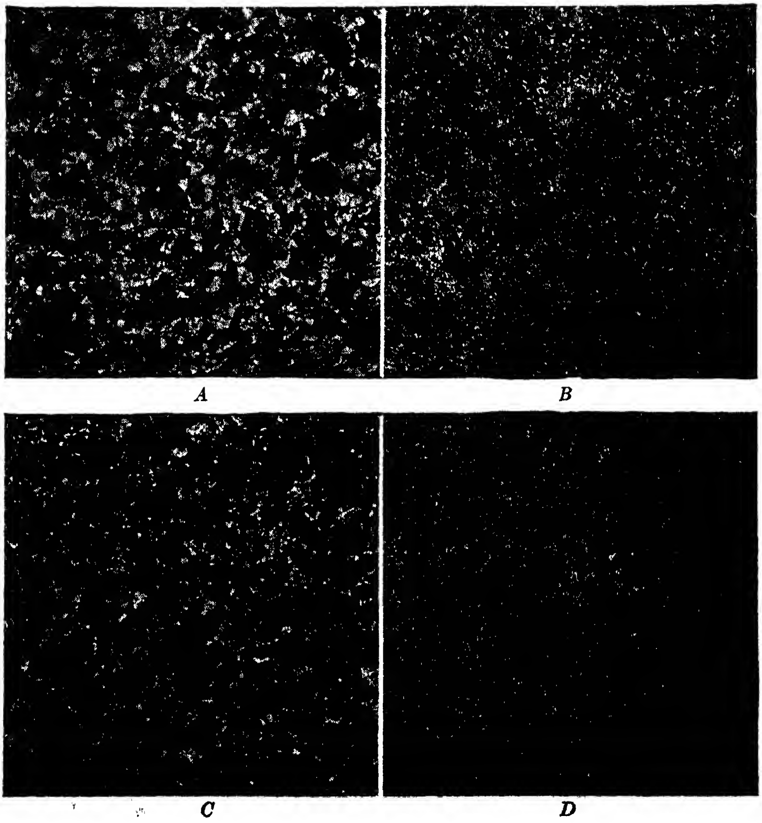

FIG. 2-XIX.—*A*, typical structure of S.A.E. 6130 steel (0.30 carbon, 0.95 chromium, 0.18 vanadium) in the annealed condition. × 100. *B*, typical structure of S.A.E. 6130 steel after water hardening and drawing at 800°F. × 100. *C*, typical structure of S.A.E. 6145 steel (0.50 carbon, 0.95 chromium, 0.20 vanadium) in the annealed condition. × 100. *D*, typical structure of S.A.E. 6145 steel after oil hardening and drawing at 800°F. × 100. (*Courtesy of Carpenter Steel Company.*)

S.A.E. 6145.—This higher carbon, oil-hardening steel is found to be very suitable for heat-treated forgings and machine parts that are subjected to severe conditions, where high strength and anti-fatigue properties are essential. It finds its principal applications in coil springs in the smaller sizes, in automobile flat-leaf springs, gears, shafts and heavy forgings, and in aircraft parts, such as studs.

In the leaf-spring application the service calls for a considerable degree of elasticity in combination with strength and hardness. The requirements are met by these steels through the hardness and strength imparted by the chromium and the toughness imparted by the vanadium.

Figure 2-XIX-*C* and *D* shows the typical structure of this steel in the annealed condition and after oil hardening and drawing at 800°F.

SILICO-MANGANESE STEELS

The effect of the combination of these two elements, as has been pointed out, is to bring about a steel with high strength and sufficient ductility for the services to which it is applied after heat-treating.

S.A.E. 9255 and 9260.—These are principally one-purpose steels, used mainly for leaf springs. They respond very readily to oil quenching and, when properly drawn, possess, in addition to the strength required by the springs, excellent toughness and resistance to shock. These steels, therefore, find wide application in automotive and railroad fields for both coil and leaf springs.

CHAPTER XX

FACTORS AFFECTING THE SELECTION OF STEELS

From the foregoing study of the so-called constructional alloy steels it is evident that with so many grades available the choice for a given application presents a considerable problem. It has frequently been mentioned that there are too many steels included in the S.A.E. specifications and that it would be desirable, if it were possible, to utilize a series of simple steels varying in carbon content according to the strength requirements. However, because of the mass of differing requirements that must be met by the various parts to which these steels are applied, no such a simplification is possible. That such a simplification, by the use of a series of plain carbon steels, heat-treated, is impossible, may be illustrated by considering the limitations of the use of such steels in machine parts. These[1] are that:

1. The part must be small enough so that it will harden completely without too drastic a quench.
2. The part must be fairly regular in cross section so that it can be heat-treated without warping beyond the possibility of minor straightening.
3. The part requires so much stiffness that it is bulky and the metal is lightly stressed.
4. The shape must be free from notches, grooves, and any other stress raisers at all critical locations.

These statements pertaining to the limitations of plain carbon steels, heat-treated, indicate—in part, at least—why it is necessary to use alloy steels; in fact, these statements define the true field of heat-treated alloy steels.

The problem of the choice among the available steels involves a consideration of such factors as cost, heat-treatment, hardenability, mechanical properties, freedom from occasional seams or excessive slag and other unsoundness, machinability, and distortion tendencies.

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[1] **Boegehold, A. L., Advantage of Alloy Steels, *Metal Progress*, 31, pp. 265–269 (1937).**

COST

When making an analysis of a given part, the metallurgist must select the steel and its heat-treatment so that when it is applied it will function satisfactorily and continue to do so for long periods of time when subjected to the rigors of the operating conditions. This result must be accomplished with the least possible cost.

If in his selection, the engineer could base his choice on the accomplishments of each alloying element, the problem would be simplified. However, since the raw-material cost of the separate elements varies widely, it is necessary to figure on the accomplishment of the resultant steel per dollar. To the increased costs for the element or elements required there must be added the extras (alloy extras) for control, the recovery efficiency in melting, which varies for each element, the smaller per cent recovery of sound metal, extra inspection, added power in fabrication, etc.

HEAT-TREATMENT

The selection of a steel for a definite application must depend not only upon the properties developed by heat-treatment but also upon the ease with which the steel may be formed and the way in which it responds to the heat-treatment generally applied in the particular practice.

The engineer has available a number of charts for the various grades of steel that supply him with data on the conservative physical properties after heat-treatment. These he may use as guides in selecting the proper steels from the standpoint of physical properties. However, after the choice is made and when the steel arrives at the user's plant, the data from the charts should be supplemented by a series of tests, so that the treatment temperatures and time at temperature can be adjusted to the particular composition, and so that there can be assurance that each lot of material will respond in a normal manner to the heat-treatment generally applied.

The appraisal of the utility of a steel and its heat-treatment for a part is in large measure dependent upon the personal judgment of the engineer. This judgment must be based upon the engineering requirements of the part and the suitability of the heat-treating facilities with respect to such factors as initial cost against the fabricating cost, hardenability with respect to section size, tendency toward distortion, machinability at various hardnesses, etc. The difficulty in choosing a given steel for an application and the variations in judgment of the suitability, as influenced by the factors just noted, is evidenced in the automotive industry, where as many as five different compositions may be employed by as many different producers for the same part.

TENSILE PROPERTIES

In making a comparison of steels on the basis of the usually determined tensile properties, the difficulties of the choice between steels is further evidenced, particularly when the basis is upon the properties developed on sizes that will completely harden, and that have been treated to develop the same tensile properties.

TABLE 1-XX

S.A.E. Number	Type of Quench
4340	Oil
4145	Oil
6145	Oil
3240	Oil
5145	Oil
4645	Oil
3145	Oil
6130	Water
5130	Water
2345	Oil
3130	Water
4130	Water
1330	Water
2330	Water

From the data that have been assembled by Janitzky and Baeyertz[1] it has been shown that there is only a small deviation from a constant relationship between the ultimate strength, yield strength, hardness, and elongation in a number of water- and oil-quenched steels when tempered to various identical hardness values. This relation is valid, provided that the steel and the

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[1] JANITZKY and BAEYERTZ, The Marked Similarity in Tensile Properties of Several Heat-treated S.A.E. Steels, A.S.M., "Metals Handbook," pp. 515–518, Cleveland, Ohio, 1939.

section size are such that the test specimen is hardened throughout when quenched or that the structure is substantially martensitic. All the data to be subsequently shown were obtained on standard 0.505 tensile specimens machined from as-treated 1-in. round bars. The steels used in the investigation are listed in Table 1-XX.

When these various steels are properly quenched so that they will harden throughout and are then tempered to the same hardnesses, the tensile properties will be found to be substantially the same. A summary of the data is shown in Fig. 1-XX.[1] The curves serve to substantiate the fact that the ultimate tensile properties may be predicted from hardness values. In this discussion of comparison of steels the curves clearly illustrate the interchangeability of the quenched-and-tempered water-hardening and oil-hardening S.A.E. steels on the basis of tensile properties, provided that the steels are quenched in such a manner that they harden throughout. It is important to note that the tempering temperature required to produce any given set of properties may vary with the different S.A.E. grades.

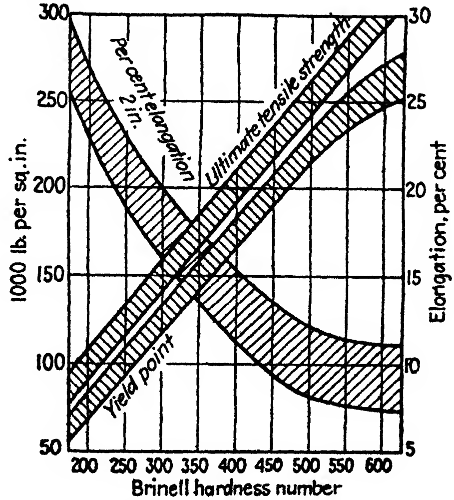

FIG. 1-XX.—The range of tensile properties in several quenched and tempered S.A.E. steels at various hardness values alike for all. (*Courtesy of E. C. Bain.*)

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[1] BAIN, E. C., "Functions of the Alloying Elements in Steel," A.S.M., Cleveland, Ohio, 1939.

When the data for reduction in cross section are plotted against hardness (Fig. 2-XX)[1] it is found that the similarity of the steels at the same strength exists only when they have been tempered below about 400 Brinell, which is equivalent to a tensile strength of about 200,000 p.s.i. According to Bain, the reduction in cross section begins to show some of the fundamental characteristics of the steels that may control notch toughness after considerable deformation has occurred. It is found that some of those steels that are able to withstand a high reduction of area prior to rupture at a given strength or at some constant high hardness value (400 to 600 Brinell) are those that allow a high tempering temperature without undue softening. The relative resistance to softening on tempering is shown in the order in which the steels used in this investigation are listed. The oil-quenched 4340 is, thus, the steel that preserves the greatest strength after a given tempering. The nickel-bearing steels, even though shown low in the order, are able to withstand a high reduction in cross section

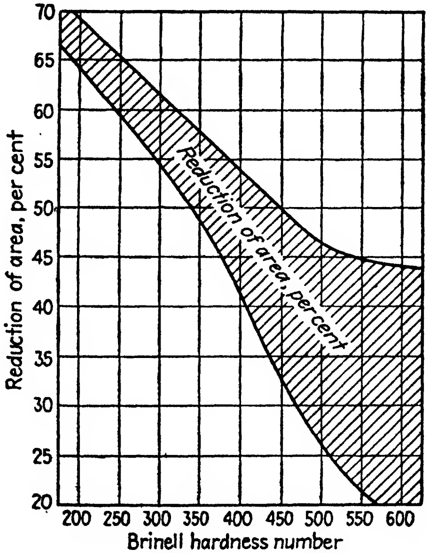

Fig. 2-XX.—The range values of reduction of areas in several quenched and tempered S.A.E. steels at a variety of hardness values alike for all. (*Courtesy of E. C. Bain.*)

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[1] Bain, E. C., "Functions of the Alloying Elements in Steel," A.S.M., Cleveland, Ohio, 1939.

prior to rupture at a given strength. It can therefore be seen that even though the tensile values are alike, the inference cannot be made that all the mechanical properties are alike. In particular, as has just been indicated, the impact toughness may be quite different.

The response to tempering is shown in Fig. 3-XX. Janitzky found that the various steels do not drop to the same hardness and tensile values when tempered at the same temperature. Instead, he found that some of them must be heated as much as 200°F. higher to reach the same values. Under the conditions imposed in the test, it was found that the tempering temperature required to obtain a given tensile strength depends on both the alloy and the carbon content of the steel and must be determined experimentally.

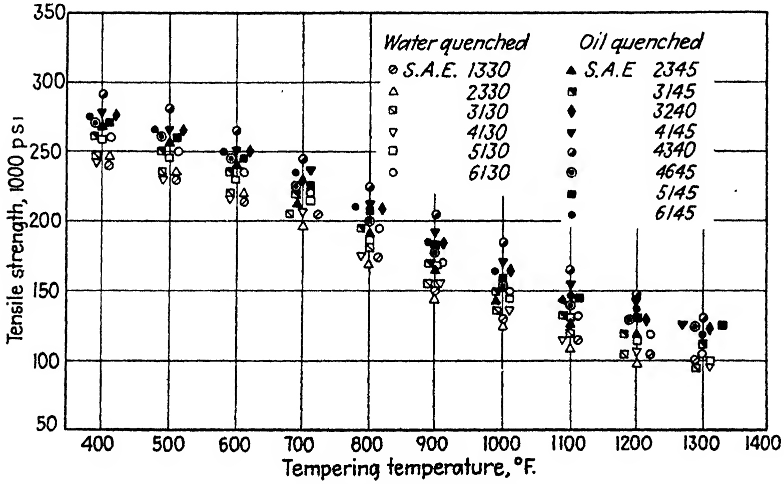

FIG. 3-XX.—The hardness in various quenched S.A.E. steels as a function of tempering temperature. (*Courtesy of E. C. Bain.*)

In order to illustrate the use of the above curves, let us assume that we are to use one of the S.A.E. steels shown in a 1-in. round section and that it is to be hardened and tempered to 150,000 p.s.i. From the curves (average values in the bands, Fig. 3-XX), the Brinell hardness is 300, the yield strength 125,000 p.s.i., the elongation 18 per cent, and the reduction of area 58 per cent. The tempering temperature needed to obtain this tensile value will vary with the steel, according to Fig. 3-XX. To obtain 150,000 p.s.i. with S.A.E. 2330 the tempering temperature will be slightly under 900°F.; with S.A.E. 6130 the temperature must be 1000°F.; and with S.A.E. 4145 it must be 1100°F.

HARDENABILITY

These figures will also help to emphasize another factor, that of hardenability with respect to section size. In all the above steels, which were treated in 1-in. sections, the depth of hardening was practically complete. In larger sections, however, the need for the use of alloys to obtain depth of hardening and high strength at the center of the section becomes evident.

In order to illustrate the relative differences in hardness penetration between various water-hardened and oil-hardened S.A.E. alloy steels, a plot (Fig. 4-XX)[1] has been made of the center hardnesses of the steels shown for different section diameters. The average center-hardness values were obtained from curves plotted from the results of transverse hardness readings taken on the different as-quenched diameters. The depth-hardness characteristics of the water-quenched S.A.E. steels are as follows: the S.A.E. 2330 and 5130 are the shallowest hardening, while S.A.E. 6130 is the deepest hardening. The plain carbon S.A.E. 1045 steel is shown (dashed on the plot) for comparison pusposes, both as a fine-grained and as a coarse-grained steel. The fine-grained steel is definitely more shallow hardening than the coarse-grained. The S.A.E. T1330, which is also shown as a dashed curve, is in the high range of the specification and, as a result, probably characterizes a higher and deeper penetration of hardness than would be normal.

The depth-hardness characterisitcs of the oil-quenched S.A.E. steel are also shown in Fig. 4-XX. In this case, the S.A.E. 3240 and 4340 are the deepest hardening; while, with the steels containing 0.45 per cent carbon, S.A.E. 3145 and 4145 are the deepest hardening, 6145 intermediate, and 2345, 4645, and 5145 are the shallowest hardening.

In the previous discussion of tensile values it has been found that there is considerable interchangeability possible with the

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[1] Klain and Lorig, Hardness Characteristics of Some Medium Carbon S.A.E. Steels, *Trans. A.S.M.*, **28**, 83–185 (1940).

S.A.E. steels on this basis. The variations in hardenability of the various steels, therefore, become a very important factor in their ultimate selection. This fact is illustrated by making comparisons between a steel of high hardenability, as S.A.E. 4145, with a steel which does not have so good a hardenability. In this case, the S.A.E. 4145 hardens better than any of the other steels; it hardens to a greater degree in all but the 3-in. sections; it holds its physical properties better as the size increases; and under any quench-and-draw treatment it will show higher physicals than a steel that does not have so good a hardenability.

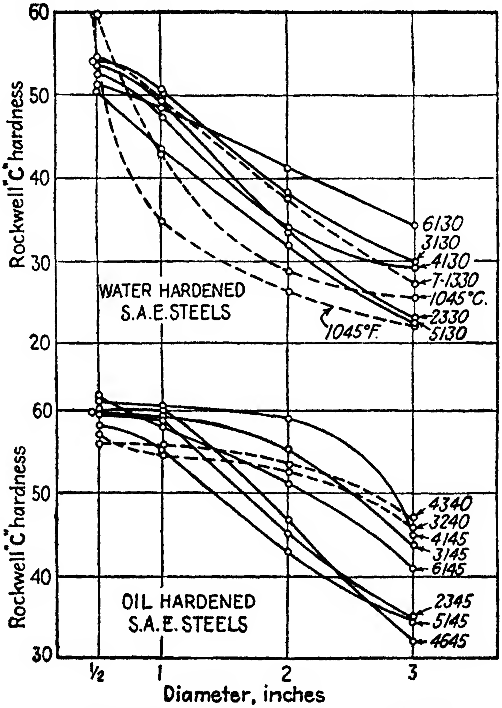

Fig. 4-XX.—Center hardness versus diameter. (*Courtesy of The American Society for Metals.*)

MACHINABILITY

Another important factor, both when comparing various grades of steel and when ultimately choosing a steel for a job, is machinability. Table 2-XX summarizes the machinability ratings for a number of steels in the cold-drawn and in the annealed condition. These figures give some indication of the relative machinability of the various steels. The machinability rating is averaged to the nearest 5 per cent from various users' estimates, and is based upon 100 per cent as the rating of A.I.S.I. steel specification B1112, cold rolled or cold drawn, when machined with a suitable cutting fluid at 180 ft. per min. under normal cutting conditions. The ratings are based on steels in a cold-drawn condition.

TABLE 2-XX.—MACHINABILITY RATINGS*

Cold-drawn				Annealed			
A.I.S.I. number	S.A.E. number	Rating	Brinell	A.I.S.I. number	S.A.E. number	Rating	Brinell
B1112	1112	100	179–229	A1330	1330	50	179–235
B1113	X1112	135	179–229	A1340	1340	45	179–235
C1118	X1315	80	143–179	A2340	2340	45	179–235
C1137	X1335	70	187–229	A3140	3140	55	187–229
C1010	1010	50	131–170	A4130	X4130	65	187–229
C1020	1020	65	137–174	A4150	4150	50	187–235
C1030	1030	65	170–212	A4640	4640	55	187–235
A2317	2315	55	174–217	A5140	5140	60	174–229
A3120	3120	60	163–207	A6145	6145	50	179–235
A4615	4615	65	174–217	E52100	52100	30	183–229
A5120	5120	65	170–212				
A6120	6120	50	179–217				

* Machinability Rating of Various Metals, *Metal Progress*, **44**, 624 (1943).

Metals that have been heat-treated will, of course, have different machinability ratings.

The Brinell hardness values shown for the various steels are intended to provide the most desirable limits for normal machinability and, in general, describe the commercial practice in filling steel purchase orders. It is generally accepted practice that the limiting value of hardness for normal machinability is approximately 355 Brinell, although many steels are machined in mass production at hardnesses of 400 Brinell or more.

It should be emphasized that the value of the relative figures given in Table 2-XX is affected by such problems as rigidity of equipment, control of speeds and feeds, the type of lubricant used, and the form of the tool. Hence, the application of these figures may not always prove satisfactory.

STRESS CONCENTRATION, DISTORTION, CRACKING, QUENCHING STRESSES

The importance of stress distribution with respect to the behavior of a steel in a given application, can be emphasized when consideration is given to the fact that stresses as applied in service are not static but vary greatly in intensity and application. It is evident from the type of fractures encountered in machine parts that the ordinary tensile test will not give satisfactory results. The fractures encountered are practically all of the type that show no elongation and occur as the result of (1) fatigue failure, failure due to a large number of stress applications above the endurance limit but below the elastic limit, and (2) breaking stresses applied at a notch or groove.

Investigations by Almen and Boegehold[1] on carburized rear-axle gears led them to the conclusion that for this service the important factor was the resistance of the steel to fatigue in overload. In the case of the alloy steels ordinarily applied, they could find little superiority of one over the other and, as a result, concluded that such factors as distortion or warping tendencies, machining characteristics, and cost should be the governing ones.

From the previously mentioned limitations of the application of carbon steels in important machine parts it is evident that carbon steels may be used only where the stress concentrations, as calculated, are low enough so that the maximum that may occur at sharp corners or grooves is well within the endurance limit. In addition, the section must be small enough so that complete hardening is obtained without resorting to drastic quenching, with its resultant danger of cracking.

It is important, therefore, in view of these factors, that the selection of steel be made so that the desired properties can be obtained without resorting to drastic forms of treatment. Water and brine quenching are considered as a drastic treatment when applied to carbon and alloy steels of over about 0.35 per cent carbon and may, when improperly applied, produce cracking or spalling, particularly at reentrant angles, sharp corners, and

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[1] ALMEN, J. O., and BOEGEHOLD, A. L., "Rear Axle Gears: Factors Which Influence Their Life." *Proc. A.S.T.M.*, Vol. 35, part 2, pp. 99–146, 1935.

changes in section. Oil quenching minimizes both distortion and the danger of cracking, but it may not give sufficient surface hardness unless the hardenability is developed by the use of alloys. Frequently, however, a soft core—hence, shallow hardening—is desirable where a warped piece must be straightened without danger of cracking. Drastic forms of treatment can be avoided in many cases by selecting a steel with a slightly higher carbon content, by the addition of alloys or possibly by an increase in alloy content, or by the control of grain size. With respect to grain size, it is found that coarse-grained steels have a greater tendency to distort than do the fine-grained; hence, grain-size control is as important, at least, as the alloy content in this regard.

POSSIBLE CHOICES

The various factors that must be considered in the choice of a given steel for a service application and the differences in judgment that arise in making this choice may be illustrated by several examples taken from discussions on steels used in various parts of passenger cars by Graves,[1] and by Boegehold.[2]

Steering Knuckle.—The study of the selection of the steel for this part will help to emphasize one of the limitations of carbon steels, *i.e.*, that when they are heat-treated to high strengths and hardness, they are unreliable if the part is irregular in shape and cross section. The steering knuckle is an odd shape having surface irregularities that cause stress concentrations. It is a very important part, which, if broken, would cause serious difficulties; hence, the material and the heat-treatment must be chosen to minimize every possibility of failure. The knuckles are in all cases forged, quenched, and tempered, and held at hardnesses ranging from Brinell 241 to 302. The types of steel utilized are T 1330 or T 1340 (high manganese), 4130 (chromium, 1 per cent), 3140 (nickel-chromium), 5130 (straight chromium), or carbon-molybdenum.

The choice of a steel for this part is based primarily upon service results. However, since the strength and toughness of

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[1] Graves, W. H., What Steel Is Best for This or That Part? *Metal Progress*, **29**, 36–42 (1936).

[2] Boegehold, A. L., Advantages of Alloy Steels, *Metal Progress*, **31**, 265–269 (1937).

the steels used are alike, the choice is ultimately made on the comparative machining characteristics, as there is considerable machining to be done after the steel is heat-treated. It is evident from the previous list that a number of steels are found satisfactory from a service angle, but as far as machinability is concerned the appraisal differs considerably and this accounts for the different alloy steels in use.

Gears.—Gears comprise a class of parts where the properties of alloy steels are very carefully used. Some of the metallurgical considerations involved in the choice of steels are as follows.

The durability of a gear of any type is dependent first of all on the engineering design and upon the manufacturing procedure. Improperly shaped teeth with inadequate fillets at the roots of the teeth will nullify all efforts to produce a long-lived gear. Similarly, in the production of the gear, the use of improperly shaped tools, causing toolmarks where there should be fillets, will nullify the efforts of good engineering design.

High surface hardness for wear resistance is required in all cases and is produced either by a cyanide dip or by heat-treating in a controlled atmosphere, which will produce the hard surface layer required. The hardness requirements of the steels used vary from Rockwell C 52 to 60, according to whether they are carburized or heat-treated. At this high hardness range and particularly where the hardness is desired throughout the part, it is important that the ductility be as high as possible, so that stress distribution may be assisted at points of stress concentration. This is one of the reasons why alloy steels are desirable in this part, since plain carbon steels, carburized, will, as far as hardness and wear resistance are concerned, be just as good as the alloy steels.

The problem of distortion occurring during hardening must be taken into consideration in all parts that are to be used without finishing operations after heat-treatment, not only because of high stress concentration on smaller areas but also because of imperfect functioning, which causes undesirable noise. In gears that are highly stressed, steels must be chosen that do not require a drastic quench to harden, even in fairly heavy sections.

The appraisal of a given steel for application in this type of part is also found to vary according to the beliefs of the different users. This is illustrated in the steels used for transmission gears, which include 4620, 4615, 4640 (nickel-molybdenum), 3145 (nickel-chromium), 6150 (chromium-vanadium), and 5135, 5145 (straight chromium).

RELATIVE USES OF OPEN-HEARTH AND ELECTRIC STEEL

In regard to the production of S.A.E. grades of steel, it is important to note that the analyses previously discussed are made in both furnaces. In general, it is found that engineering design and regulation of heat-treatment in the automotive industry is developed at the present time to the point where open-hearth steels are entirely satisfactory. In some cases, however, owing to certain limitations in design or weight, it is desirable to use a higher quality of steel. Examples of these cases are found in heavy-duty truck gears and axles and in certain rock-drill parts made from alloy steels, where considerable quantities of electric-furnace steel are being applied. Because of the severe fatigue service encountered by these parts, it is absolutely necessary to have clean, uniform steel, and electric-furnace steel is quite generally used. The choice of electric-furnace steel, therefore, results from certain unusual service requirements where clean steel and uniform steel, from the standpoint of grain size, hardenability, and analysis limits, are necessary.

THE NATIONAL EMERGENCY STEELS

To meet the conditions of the Second World War, the selection of steels has been somewhat simplified. This simplification is the direct result of real problems, which necessitated (1) the simplification and streamlining of analyses, to facilitate maximum production from melting equipment; (2) the economic use of alloying elements, to effect maximum results from a minimum supply; and (3) the utilization of valuable alloys contained in an undesirable form in contaminated scrap. It has resulted in the development of newer and leaner alloy steels, which are proving in many cases superior to the older and richer alloy steels. These new alloys are known as the NE, or National Emergency, steel compositions.

Conditions surrounding the production of alloy steels have made it necessary that periodic changes be made in the chemical composition of the NE steels, but in general the steels of the major NE series are as follows:[1]

1. NE 1300 series, carbon-manganese steels.
2. NE 8600 series, nickel-chromium-molybdenum steels.
3. NE 9200 series, silicon-manganese and silicon-manganese-chromium steels.
4. NE 9400 and 9500 series, manganese-nickel-chromium-molybdenum steels.
5. NE 52100, carbon-chromium steels.

These steels, owing to the urgent need for their development, were manufactured and compared on the basis of hardenability tests of various combinations of predetermined compositions containing the elements shown above. As a result of these tests, new steels have been devised that have hardenability values comparable to the previously used S.A.E. steels. At the present time and until more precise data can be gathered on their physical properties, these steels can be compared and substituted only on the basis of comparable hardenabilities. This comparison and resultant substitution are proving entirely satisfactory and it is felt that the ultimate future results will be the retention of this simplified system in lieu of the past, complicated system, which involved approximately 4,000 compositions.

Suggested Questions for Study and Class Discussion

Table 3-XX shows the comparative properties between casehardening and hard-tempering alloys of the same alloy content, the difference in the two steels being in the carbon content—the casehardening steel having low carbon and the other above 0.50 per cent carbon for tempering.

From Table 3-XX and from your knowledge of the effects of the various alloy additions on steels, explain what steel you would choose in making the following products:

1. Truck or passenger-car transmission gears.
2. Truck or passenger-car ring or pinion gears.
3. Mesh gears in a lathe.
4. Piston pins.
5. Automobile axles, both front and rear.
6. Automobile flat-leaf springs.
7. High-strength, lightweight wrenches.
8. High-strength, low-weight, adequately corrosive-resistant material for the construction of railway and traction cars, trucks, and buses etc.

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[1] For actual compositions see "The National Emergency Steels," American Iron and Steel Institute, New York, and "The National Emergency Steels," A.S.M., Cleveland, Ohio.

In formulating your answers concerning the choice of steels for these applications, consider the following points:

1. What types of service conditions are encountered in such applications?

TABLE 3-XX*

Cost and properties	Steel	
	Case hardening alloys	Hard-tempering alloys
Initial cost of steel	Same	Same
Cost of forging	Less	Greater
Cost of annealing forgings	Less	Greater
Cost of machining	Less	Greater
Cost of finish heat-treatment	Greater	Less
Distortion in heat-treating	Greater	Less
Ease of straightening	Easier	More difficult
Hardness of surface	Greater	Less
Wear resistance	Greater	Less
Toughness of surface	Less	Greater
Tendency to pit under high unit pressures	Less	Greater
Tendency to upset under heavy loads	Greater	Less
Over-all strength of part	Less	Greater

* "Alloy Steels Simplified," The Carpenter Steel Company, Reading, Pa.

2. Does this application demand deep hardness penetration? High strength or toughness? Wear resistance? Minimum amount of distortion and cracking in hardening? Etc.

3. Into which group of steels do these requirements fall?

a. Low carbon casehardening.

b. Medium carbon tough tempering.

c. High carbon hard tempering.

4. What alloy steels in this group will best meet such service requirements?

CHAPTER XXI

SPECIAL ALLOY STEELS

This important class of steels, designated as the *special alloy steels*, includes those to which alloy additions have been made, usually in quantities exceeding the additions made in the previously discussed class. Such additions bring about very valuable characteristics, other than merely improved mechanical strength. It should be noted that these steels, in general, are employed for purposes where ordinary steels will not serve satisfactorily.

The following important classes of steels will be studied: stainless and rust-resisting steels, acid-resisting steels, heat-resisting steels, wear-resisting steels, tool steels.

CORROSION- AND HEAT-RESISTING STEELS

In the general discussion of this group of steels the word "stainless" will be frequently used, although it is a rather abused term in this application, since practically any ferrous material will stain under certain conditions. Therefore, since "stainless" is but poorly descriptive of this type, the expression "corrosion- and heat-resisting" steels seems preferable.

The uses to which this group of steels may be applied are unlimited. It was first generally used in this country for cutlery but, during the last few years, the largest tonnages have been devoted to decorative purposes. The widest and most valuable applications of this steel, however, are in the industries, for such things as bakery equipment, chemical-plant equipment, dairy equipment, furnace construction, paper-manufacturing equipment, petroleum refining, turbine parts, pump rods, liners, etc. Indeed, the applications of corrosion-resisting steels are and will be limited only by their cost.

No one metal or alloy is equally adaptable to all purposes, although the development of the alloys of chromium and iron has made available, for industrial purposes, a group that approaches complete resistance to atmospheric corrosion, with very desirable physical properties. Chemical and metallurgical tests indicate that approximately 11 per cent of chromium is the minimum amount necessary to provide this characteristic resistance.

The fact that chromium, when combined with iron or low carbon steel, imparts noncorrodible properties, is probably due to the formation of a characteristic, very thin, surface film of oxide on the clean metal. This is conceived as being continuous, stable, so thin as to be invisible, yet an effective barrier against the further action of corrosive agents. At elevated temperatures this film is equally important. It becomes thicker, loses its transparency, and eventually forms a definite scale, which is continuous and firmly adherent and serves to protect the underlying surface from further oxidation. It is the formation of this scale that provides the oxidation-resisting qualities of these alloys at high temperature.

The straight chromium steels are usually considered ferritic and their susceptibility to hardening depends upon the chromium and carbon contents. Increasing the carbon or decreasing the chromium makes them more susceptible to hardening.

The addition of substantial amounts of nickel to the chromium-iron system provides a series of alloys having the property of greater resistance to corrosion and at the same time introduces very definite physical properties.

Nickel also has the property of developing surface passivity in a manner similar to chromium, except that somewhat larger percentages of nickel are necessary to obtain similar results. Structurally, nickel has a very marked effect on the chromium-iron alloys. Whereas chromium and iron form solid solutions that are normally in the alpha (or ferritic) condition, the introduction of substantial amounts of nickel causes the alloys to assume and retain the gamma (or austenitic) state, provided that they are properly processed and heat-treated.

These austenitic alloys are nonmagnetic and are not hardenable by heat-treatment. They have high ductility, as properly annealed, and are well adaptable to deep-drawing operations. Cold working causes a rapid increase in the strength and hardness of these alloys. Advantage is taken of this property for the production of cold-rolled strip, wire, and tubes of very high strength.

In the space allotted it is impossible to discuss the great number of these steels now available. This discussion will concern specifically a few of the different types of corrosion- and heat-resisting steels, with the outstanding characteristics of a few of the modified types.

Five Per Cent Chromium Steels.—This group of low carbon steels containing 4 to 6 per cent chromium is obviously not one of the "stainless" steels. However, from the viewpoint of heat and corrosion resistance, such analyses do fall into this section. Further, it is well beyond the range of heat-treated alloy steels of the S.A.E. types.

This group, along with its modifications, finds excellent application in tubing, castings, and other equipment for moderately severe service involving temperature, pressure, and corrosion—particularly in oil refineries, where the service is too severe for plain carbon steels. For more severe service the higher chromium or high chromium-nickel alloys should be applied.

Commercially, this group of chromium steels is now produced with chromium contents ranging from 2 to 9 per cent, or higher, and carbon from 0.10 to 0.20 per cent; the composition varies with the severity of the service conditions.

TABLE 1-XXI*

Carbon, per cent	Chromium, per cent	Air cool, °F.	Draw	Yield point, p.s.i.	Tensile strength, p.s.i.	Elongation, in 2 in., per cent	Reduction, per cent	Brinell hardness	Izod impact, ft.-lb.
0.20	5.20	1575	None	114,000	212,310	9.0	18.5	417	23.0
0.20	5.20	1575	1110°F.	120,000	137,100	18.5	58.7	272	36.0
0.20	5.20	1575	Furnace cooled	33,200	75,600	32.0	75.0	152	84.1

* WRIGHT, E. C., Rolled Materials, Especially Tubing, in "The Book of Stainless Steel," 2d ed., pp. 234–244, E. E. Thum, ed., A.S.M., Cleveland, Ohio, 1935.

Heat-treatment and Physical Properties.—The most noticeable property of this group of steels is that they have intense air-hardening properties, proportional to both the carbon and the chromium contents. This property may best be seen in Table 1-XXI.

These figures, although they represent only one carbon and chromium content, show that these steels have high yield points and tensile strengths by simple air hardening, or air hardening and drawing—treatments that are as effective as the quenching of ordinary steels in a liquid medium. Combined with these, they have high ductility and impact toughness that is superior in many cases to some of the best structural alloy steels, heat-treated to the same tensile strength. A rather peculiar physical property is the low elastic ratio in the annealed state, which is often less than 50 per cent, whereas the air-hardened and tempered steels show a very high ratio, usually over 80 per cent.

Carbon will, of course, have an appreciable effect on the yield point and tensile strength, especially on the annealed material. In applications where the alloy is to be used for high-temperature service, such as cracking still tubes and condenser tubes, and where flanging, rolling, and other manipulations are required, only the soft-annealed material should be used.

Corrosion Resistance.—This group of alloys, according to investigations,[1] shows good resistance to attack by such solutions as aerated Pittsburgh city water at 140°F.; aerated hydrochloric acid, sodium chloride, dilute sulfuric acid, all at 70°F.; and sodium chloride aerated with hydrogen sulfide at 70°F. The chromium addition decreased the corrosion damage to one-half or less. These alloys resisted high sulfur oils and hot hydrogen sulfide between three to ten times as well as would regular steel.

Structure.—Being familiar with the structural diagram of the 6 per cent chromium steels, we can more or less postulate the type of structure that can be obtained following annealing and air cooling. The structure of a rolled or forged bar (of the 5 per cent chromium alloys) after annealing or furnace cooling from 875°C. (1610°F.) is shown in Fig. 1-XXI-*A* and *B*. The carbides in the steel are in a globular or coalesced condition, dispersed throughout the ferrite matrix, instead of having the usual platelet structure.

In air cooling from 875°C. (1610°F.) the effect of chromium in lowering the reaction rate of gamma to carbide becomes evident; the structure is fine and the resultant physical properties are greatly increased. The actual result is comparable to the effect of the liquid quenching of high-strength construction steels.

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[1] WRIGHT, E. C., Rolled Materials, Especially Tubing, in "The Book of Stainless Steel," 2d ed., pp. 234–244, E. E. Thum, ed., A.S.M., Cleveland, Ohio, 1935.

Figure 2-XXI-*A* and *B* shows the fine dispersion of the carbides throughout the ferrite matrix.

The stability of structure is of considerable importance when the material is exposed for long periods at high temperatures. A change in the metal structure during such service may change the volume and dimensions, reduce the strength and ductility, and alter its resistance to corrosion. It is very important, then, that we know what structural transformations take place and what are their resultant effects. It is found that the effect of long-time, high-temperature service is similar to annealing, since these changes, which occur after long periods of time at tempera-

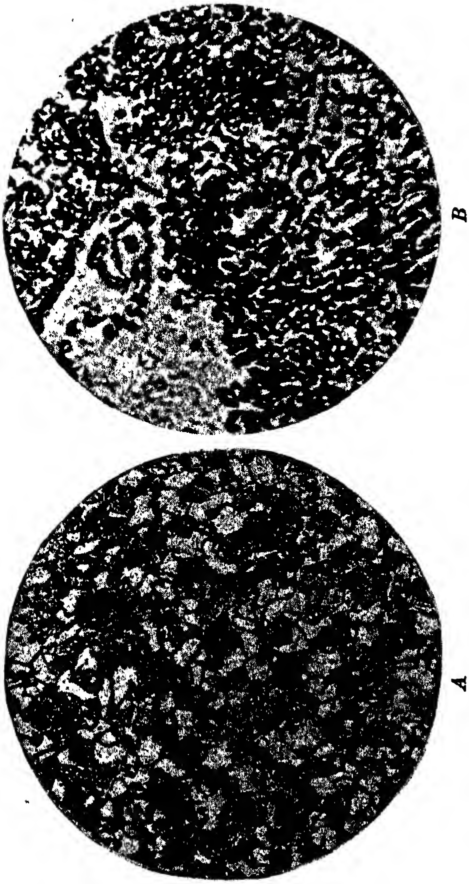

FIG. 1-XXI.—Structure of 5 per cent chromium steel, after furnace cooling from 875°C. (1610°F.) etched in 4 per cent nitric acid. *A*, 100 diameter. *B*, 1000 diameters. (*Courtesy of American Society for Metals.*)

tures of 900 to 1400°F., are the same as those that take place after annealing in the ordinary critical ranges of most steels, *i.e.*, spheroidizing of the carbides. It, therefore, follows that a suitable preheat-treatment to place the metal in equilibrium with operating conditions, inhibiting thereby further structural changes, would be to anneal fully before installing. Since the 5 per cent chromium steels assume the most stable arrangements of the carbides (globular) after furnace annealing, it can be assumed that there will be no further rearrangement of constituents upon use.

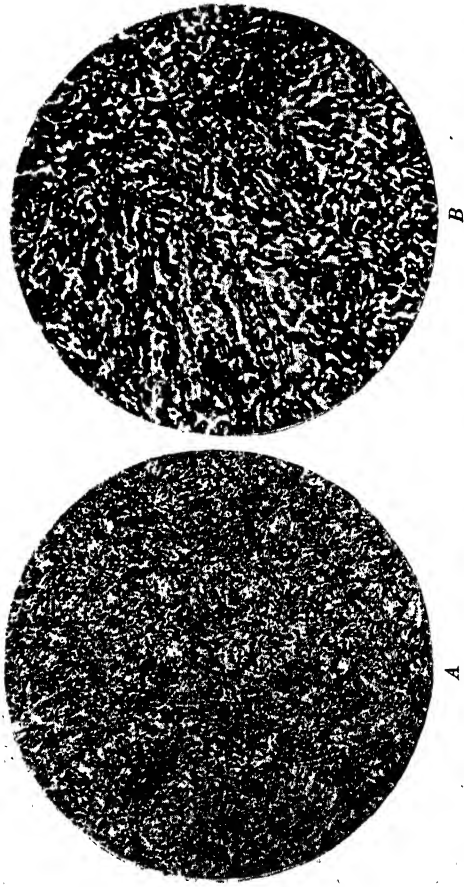

FIG. 2-XXI.—Structure of 5 per cent chromium steel, after air cooling from 875°C. (1610°F.) etched in 4 per cent nitric acid. *A*, 100 diameters. *B*, 1000 diameters. (*Courtesy of American Society for Metals.*)

Modifications of the 5 *Per Cent Chromium Alloys.*—The present tendency in the development of this group of alloys is toward the addition of such elements as molybdenum, tungsten, silicon, and titanium with the chromium.

Titanium, in amounts around 0.75 per cent, is the only element of those named above that, when added to the chromium steels, affects the corrosion resistance, and it is only moderate in its action. It apparently accomplishes this effect by combining with the carbon to form titanium carbides and leaves the chromium, which would ordinarily be in combination with the carbon, in solid solution in the iron to impart its valuable corrosion- and scale-resistant characteristics.

Tungsten, molybdenum, silicon, and titanium all affect the oxidation resistance of chromium steels. From 1,000-hr. oxidation tests at 1200°F., it is found that a 1 per cent tungsten addition increases the resistance about one-fifth, while 0.50 per cent molybdenum lowers the resistance about the same amount. Silicon increases the oxidation resistance of both plain carbon and alloy steels. This element alone offers sufficient protection from serious oxidation up to 1200°F. By use of suitable combinations of both silicon and chromium, a number of compositions have been developed that have high resistance to oxidation up to 1300°F. In these alloys the silicon may be regarded as a substitute for chromium as far as oxidation resistance is concerned. Titanium has also been observed to increase the oxidation resistance of the 5 per cent chromium steels. This improvement is due partly to the adherent scale that is formed and partly to the more favorable distribution of chromium.

Absence of brittleness is one of the first requisites for high-temperature service. This means that the steel must be initially tough at room temperature, retain good shock resistance at operating temperatures, and not become brittle upon cooling to room temperature after long heating in service. "Tensile brittleness" at elevated temperatures, which is characterized by sudden rupture of the metal without appreciable plastic deformation, is also detrimental, since a tube under pressure may fail without the bulging that would serve as a warning against its continued use.

The plain chromium, chromium-silicon, and chromium-tungsten steels are all susceptible to temper embrittlement during service. Molybdenum steels, on the other hand, are not susceptible, about 0.50 per cent being effective in eliminating the temper brittleness of 5 per cent chromium steels.

Large amounts of silicon may cause grain growth, with its accompanying loss of shock resistance. Silicon should always be used in the presence of sufficient carbide-forming element, such as chromium, manganese, and molybdenum, to restrict this tendency.

Titanium is added to 5 per cent chromium steels primarily to prevent air hardening (nonhardenable), as well as to serve in its functions of increasing the corrosion and oxidation resistance.

The creep rate of low alloy steels at elevated temperatures can be retarded by such elements as molybdenum, tungsten, vanadium, manganese, and chromium. Molybdenum is the most powerful element in raising the creep strength of steels, especially at temperatures around 1000°F. One per cent seems to be about the approximate economic limit in the unalloyed or low alloy steels, as increased amounts are not reflected in a proportional increase in creep strength. Tungsten and vanadium also improve the creep strength in the range of 800 to 1200°F., but the cost of adding enough to produce the desired effect is often prohibitive. Manganese and chromium also improve the creep strength but are much less effectual than the other elements named.

In the modifications of the 5 per cent chromium steels and steels in general, the elements discussed may be added in various combinations to produce comparatively inexpensive alloys capable of satisfying many of the service requirements discussed. The principal functions of these elements are that silicon in such compositions would furnish the oxidation resistance; chromium, the corrosion resistance; and molybdenum, the toughness and creep resistance at elevated temperatures.

Twelve Per Cent Chromium Steels.—This straight high chromium steel in low carbon grades is selected for applications wherein high physical properties are necessary and corrosion-resisting requirements are not the most exacting. Its most notable applications are for turbine blades, coal screens, valve stems and seats, pump parts, bolts and nuts, etc.

This particular grade of "stainless" steel or "stainless" iron usually contains 0.08 to 0.12 per cent of carbon, with the chromium content ranging from 11.5 to 13 per cent. Steels within this specification are both heat-treatable and corrosion resistant.

Heat-treatment.—Heating this grade for forging takes about twice as long as for ordinary steels, the operation consisting of a preheat to 1600°F. followed by more rapid heating to 1900 to 2000°F. Forging may then be effected without injury down to about 1600°F., after which reheating is recommended. In order that the metal may be machined without a special anneal, it should be cooled very slowly from the forging heat.

TABLE 2-XXI*

Properties	Annealed	Quenched 1700–1750°F., drawn 1100°F.
Ultimate strength, p.s.i.	65,000	125,000
Yield point, p.s.i.	35,000	100,000
Elongation in 2 in., per cent	35	20
Reduction of area, per cent	65	60
Impact strength, ft.-lb	80 (Izod)	75 (Charpy)
Brinell hardness	140	230
Stress causing 1 per cent "creep" in 1,000 hr.:		
At 1000°F.	....	13,000
At 1200°F.	....	2,300
At 1350°F.	....	1,400
Scaling temperature 1300°F.	....	

* WRIGHT, E. C., Rolled Materials, Especially Tubing, "The Book of Stainless Steel," 2d ed., pp. 234–244, E. E. Thum, ed., A.S.M., Cleveland, Ohio, 1935.

For complete softening, the normally cooled or cold-worked metal is preferably heated for a given length of time at 1450 to 1500°F.

Even though this steel is quite low in carbon, it can be heat-treated or hardened in much the same way as all ordinary steels, except that it does not develop the great hardness and brittleness of tool steels. The best hardening procedure is to heat to a range of approximately 1750 to 1850°F. and to follow this by a rapid air cool for thin sections and an oil quench for larger sections. Some applications, particularly springs, are made with the steel in the as-quenched condition, but generally it is desirable to sacrifice some of the resulting strength by tempering. It is desirable to avoid tempering between 800 and 1100°F. as there is a marked drop in impact values and corrosion resistance in heating within this range. Adequate impact strength can be secured by the proper tempering practice without reducing the corrosion resistance.

Table 2-XXI shows the typical physical properties after annealing and after quenching and drawing.

Structural Characteristics.—Investigation of a range of chromium steels with low carbon content has shown that, up to 13 per cent chromium, the alloys react much the same in heat-treatment, while above 13 per cent there is a marked dropping off of the physical properties, so that with very high chromium (around 18 per cent) the hardening effect by quenching is entirely suppressed. In forging the various alloys, it is found that those under 13 per cent chromium hardened on air cooling and had a fine, dense structure. Those over 14 per cent chromium appeared much softer. They did not harden to the same extent on cooling; and they showed a serious tendency toward grain growth.

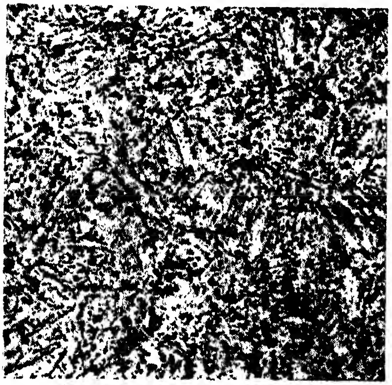

FIG. 3-XXI.—Structure of 12.25 per cent chromium, 0.08 per cent carbon alloy, after oil quenching from 1850°F. and drawing at 1175°F. etched in aqua regia, Brinell 228. × 200. (*Courtesy of Latrobe Electric Steel Company.*)

With low carbon content, the maximum response to heat-treatment is obtained when the chromium is held to a maximum of 13 per cent. This is due to the fact that the tip of the gamma loop is approached (Fig. 6-XVI) and, with more chromium, the steels become ferritic and no longer will be affected by heat-treatment. A study of the microstructure will serve to explain this phenomenon and also to show the typical structures of this class.

Figure 3-XXI shows the structure obtained in quenching and drawing alloys under 13 per cent chromium. This steel, which contained 12.25 per cent chromium and 0.08 per cent carbon, has been oil quenched from 1850°F. and drawn at 1175°F.; it shows

the uniform structure (probably sorbitic) obtained by this treatment. Quenching these alloys from 1750°F. or air cooling from the forging heats gives a uniform martensitic-like structure. Repeated heatings to 1950°F. did not alter their structural conditions.

Figure 4-XXI-*A*, *B*, and *C* shows an alloy above 13 per cent, a 14.26 per cent chromium, 0.08 per cent carbon alloy. *A* of the figure shows the forged structure, which was air cooled following forging. The structure here is composed of two distinguishable constituents, *i.e.*, ferrite (irregular white) and martensite (dark). *B* is the forged structure annealed at 1500°F., and *C* is the structure following a quench from 1750°F. and a draw at 1000°F. In both structures the only change that has occurred is the globularization of the carbides. The grains apparently have been little changed by these treatments; but should this alloy be heated to a high forge temperature or be subjected to numerous low-temperature heatings, grain growth will take place. Such large grains are entirely stable and no heat-treatment will alter them.

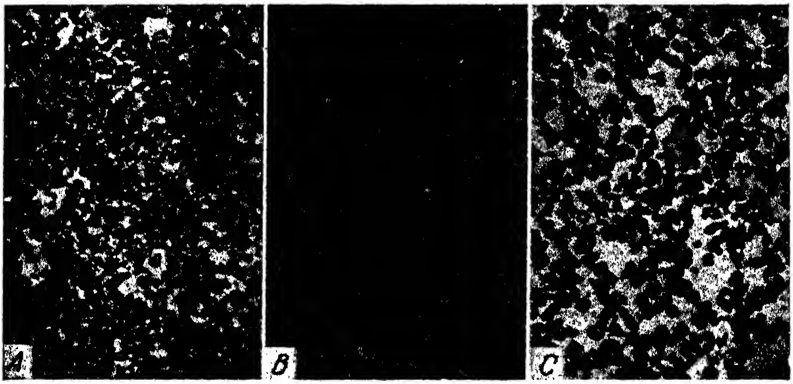

FIG. 4-XXI.—Structure of 14.26 per cent chromium, 0.08 per cent carbon alloy. × 100. *A* is the forged structure, air cooled after forging; *B* is forging annealed at 1500°F.; *C* is after oil-quenching from 1750°F. and drawing at 1000°F. (*Courtesy of American Society for Metals.*)

Mechanical Properties.—The general effect of tempering on the mechanical properties of oil-quenched alloys of this group is illustrated in Fig. 5-XXI.[1] The outstanding feature of these

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[1] "Handbook on Stainless Steels and Irons," Latrobe Electric Steel Company, Latrobe, Pa., 1937.

curves is the sharp reduction in the tensile strength, yield strength, and hardness over a relatively narrow range of tempering temperatures (800 to 1000°F.) Within this particular range there will also be a lowering of the impact values, which indicates an embrittlement in the steel. This, according to Giles,[1] is associated with the breaking up of the martensite and the formation of troostite. This intermediate product between martensite and granular pearlite shows its most pronounced influence in the low chromium group. Within this particular narrow range it is difficult to secure reproducible results, hence the steels are generally tempered at around 1175 to 1200°F.

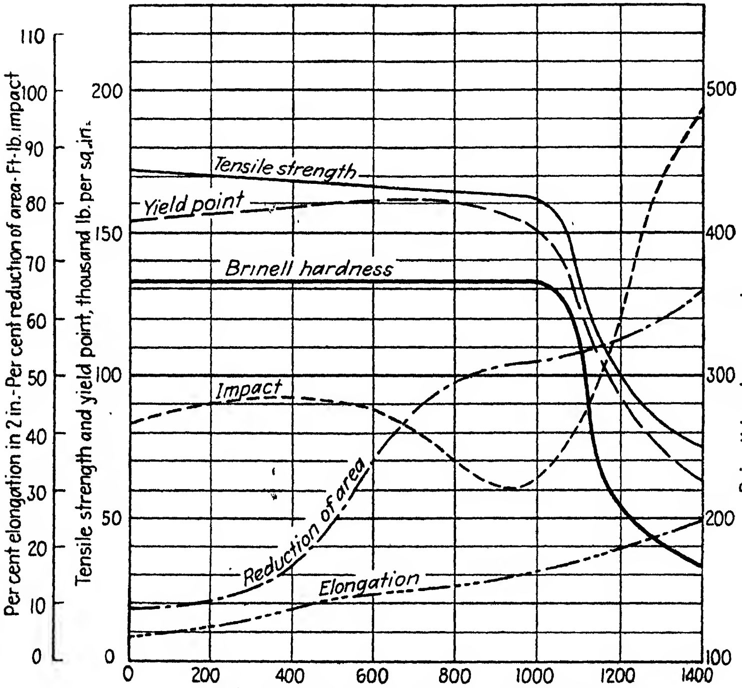

FIG. 5-XXI.—Physical property chart for a "stainless" iron containing 0.09 per cent carbon and 12.21 per cent chromium.

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[1] GILES, D. J., Sheet and Strip, "Book of Stainless Steels," 2d ed., pp. 267–276, E. E. Thum, ed., A.S.M., Cleveland, Ohio, 1935.

Cutlery Type.—This type of steel from a chemical standpoint is logically placed in this sequence, but from a development standpoint it should be discussed first. The so-called *stainless cutlery steels* belong to the first stainless alloy to be widely publicized and are regarded as being the historical beginning of the stainless group.

The original cutlery type, which contains about 0.35 per cent carbon with 13.5 per cent chromium, has very important applications in automotive parts, where wear is a factor, in ball bearings, high-grade cutlery, dental instruments, surgical instruments, manicure instruments, needles, scissors, and many similar applications.

The modified cutlery type, which contains about 0.70 per cent carbon with 16.50 per cent chromium, is an endeavor to incorporate some of the advantages of carbon steel into stainless without altering its corrosion resistance. In most respects, it is similar to the standard cutlery type, with the exception that it has a greater hardness and makes possible a better cutting edge. It has its applications in cutlery, dental instruments, surgical instruments, balls, ball bearings, valve seats, machine parts, and many others where corrosion resistance, combined with a good cutting edge or a hard bearing surface, is required.

Heat-treatment.—In all the operations it is found necessary to preheat the stock for a considerable time in the softening range and also to hold it at the required heat longer than carbon steels. Both of these steels are air hardening, so that the forgings will be hard after cooling. The hardness will depend upon the composition, the temperature to which it had previously been heated, and the cross section of the piece.

Annealing.—Stresses that have been induced by forging or machining in either of the two types can be relieved by holding the steel at a temperature of around 1300°F. for a short time and then cooling in air. In annealing for extreme softness the steel should be held for the period of time in proportion to its mass at the temperatures given in Table 3-XXI and then slowly cooled in the furnace. When annealed at this temperature, the steel is generally difficult to machine, being "gummy," which induces dragging and tearing of the material. Good machining properties may be brought about by heating at a temperature of 1400 to 1450°F. for several hours and then cooling either in the furnace or in the open air. Increased machinability may also be obtained by addition of 0.00 to 0.30 per cent sulfur or selenium.

Hardening.—Extreme caution must be taken in heating these steels to the hardening temperature as they absorb heat quite slowly, owing to their low thermal conductivity. They should be thoroughly preheated to about 1450°F. before heating to the hardening temperature and then must be held in the furnace from two to three times as long as ordinary steels to reach a uniform temperature.

Table 3-XXI*

Operation	Temperature range, °F.	Hardness
Original Cutlery Type		
Forging	2000–1700	
Softening for machining	1400–1450	Brinell 220
Annealing for softness	1575–1625	Brinell 165
Preheating	1450–1500	
Hardening	1775–1850	Brinell 550–R.C.–56
Tempering	350– 900	R.C. 51–56
Modified Cutlery Type		
Forging	2000–1700	
Softening for machining	1400–1450	Brinell 230
Annealing for softness	1625–1675	Brinell 200
Preheating	1450–1500	
Hardening	1825–1925	Brinell 600–R.C.–60
Tempering	300– 800	R.C. 54–60

* Parmiter, O. K., "The Book of Stainless Steels," 2d ed., pp. 278–292, A.S.M., Cleveland, Ohio, 1935.

The proper hardening range for the two types can be obtained from Table 3-XXI. It is found to depend upon the variations in composition, the size of the section, and the hardness desired. Since the steels are air hardening, thin sections or intricate shapes that are likely to warp or break if quenched can be air-cooled from the higher temperatures of the range. For all general purposes a quench in oil from 1800 to 1850°F. will give best results. The higher the temperature within reasonable limits, the better will be the general resistance to strain.

Tempering.—The properly quenched and tempered steels of these two types will show very high tensile strength. The extent to which the steel is drawn is limited by the purpose for which it is to be used.

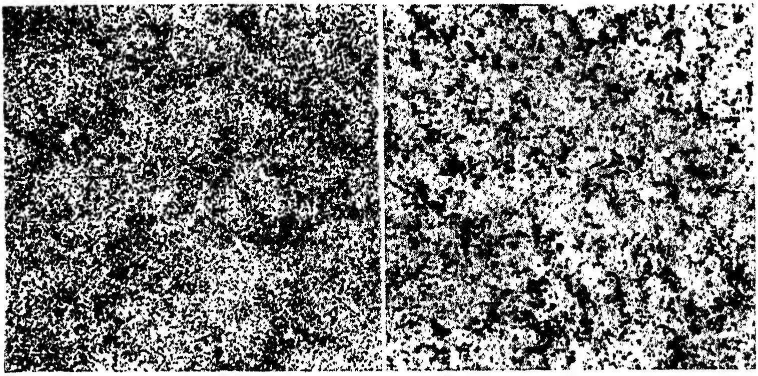

FIG. 6-XXI.—Structure of a 13.70 per cent chromium, 0.35 per cent carbon alloy. *A*, in the annealed condition. Brinell 207. Etched with aqua regia. × 200. *B*, after oil quenching from 1825°F. and a draw at 550°F. Rockwell C-51. Etched with aqua regia. × 200. (*Courtesy of Latrobe Electric Steel Company.*)

Structure.—From our knowledge of the effect of chromium additions and carbon additions to these steels, we should expect that the 13.5 per cent chromium 0.35 per cent carbon type would, on annealing, be of the pearlitic type with a slight excess of carbides, the carbides, owing to the annealing, being in globular form. Hardening this steel should produce a martensitic structure in which the carbides are held in solution. Tempering should produce the usual precipitation of the carbides. Figure 6-XXI-*A* shows the annealed structure of a

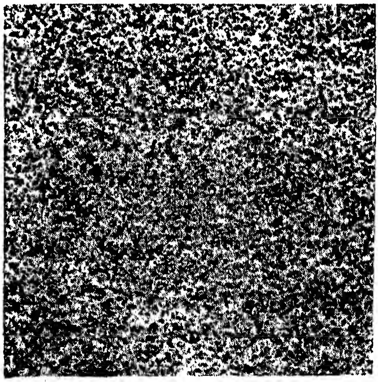

FIG. 7-XXI.—Structure of a 16.69 per cent chromium, 0.71 per cent carbon alloy after oil-quenching from 1850°F. and a draw at 600°F. Rockwell C-55. Etched with aqua regia. × 200. (*Courtesy of Latrobe Electric Steel Company.*)

13.70 per cent chromium 0.35 per cent carbon steel, while *B* shows the structure after oil quenching from 1825°F. and a draw at 550°F.

The modified cutlery type would be equivalent to a very high carbon steel. In the annealed condition the steels would be ferritic plus a considerable amount of excess carbides in the globular condition. Quenching should produce a martensitic matrix with considerable excess of carbides still existing. Figure 7-XXI shows the structure of a 16.69 per cent chromium 0.70 per cent carbon steel after oil quenching from 1850°F. and a draw at 600°F.

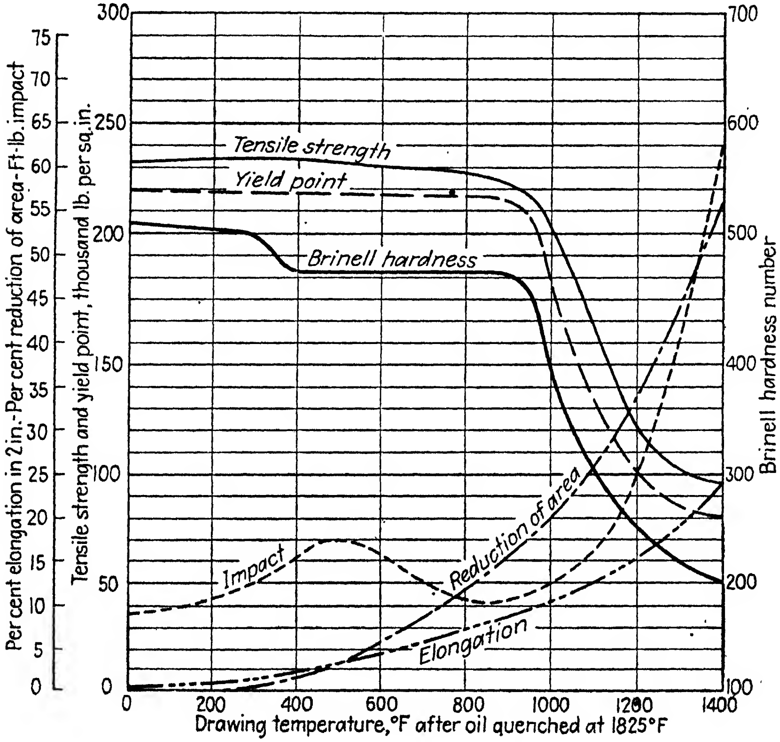

FIG. 8-XXI[1].—Physical property chart for a "stainless" steel containing 0.34 per cent carbon and 13.55 per cent chromium.

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[1] "Handbook on Stainless Steels and Irons," Latrobe Electric Steel Company, Latrobe, Pa., 1937.

Properties.—Probably the most outstanding feature of these grades of steel is their surface stability or resistance to stain, corrosion, and rust. In the lower chromium-carbon grades these properties are developed to a maximum degree only when in the hardened condition or some modification of it, and when a clean, smooth surface is exposed. This is due to the fact that in the hardened condition all the carbides are in solution, leaving the chromium to effect its corrosion-resisting property. This property will persist even after tempering to 900°F.

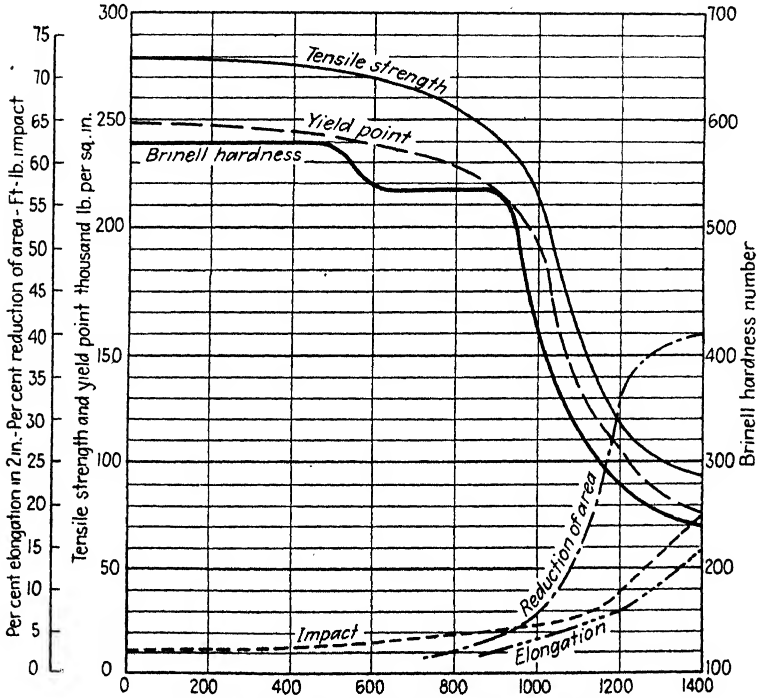

FIG. 9-XXI[1].—Physical property chart for a "stainless" steel containing 0.65 per cent carbon and 17.30 per cent chromium.

The higher carbon-chromium types of stainless steels have several decided advantages over the previously described type. The first advantage is that, after suitable heat-treatment, they are decidedly superior in their corrosion-resisting properties.

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[1] See opposite page for footnote.

The second is that they will attain a much higher degree of hardness. Thus, where great hardness, high tensile strength, and high corrosion resistance are desired, they are particularly adaptable.

Physical Properties.—A comparison of the physical properties of the two types of steels discussed may be made from Figs. 8-XXI and 9-XXI.

Fabricating.—In the annealed condition these steels may be cold-worked or drawn only to a limited amount. Stresses set up by this work are inherent in the steel and must be removed by heat-treatment. Cold working also puts the material in a state susceptible to stain. All cold-worked material should be annealed and hardened in the same manner as the hot-rolled steel, and then ground and polished, to obtain full advantage of its stain-resisting properties.

All the better quality stainless-steel cutlery is forged from round stock in order to form solid, one-piece bolsters. In the cases where a fairly good quality of knife is required, the original cutlery type or modified cutlery type is applied. In the lower quality types of knives and forks, production methods are used, the common composition being known as stainless iron (0.12 per cent carbon, 12.5 per cent chromium). In any event, the steel is hot-rolled into the desired double bevel or flat size and carefully annealed. From these sections the paring-knife, table-knife, and bread- or carving-knife shapes are produced. These blanks are then heat-treated to the required hardness and resiliency for a satisfactory knife, after which they are carefully ground, polished, and mounted in handles with special bolsters.

Seventeen Per Cent Chromium Iron.—Possibly the largest tonnage of straight chromium steels is made from this type. It is entirely permanent in ordinary atmospheres, provided that its surface is well polished and free from foreign material. It is a nonhardenable alloy, which develops and retains its full corrosion resistance without any special heat-treatment. It is quite ductile and malleable and is suitable for all except the most drastic drawing operations. It has a high resistance to oxidation at elevated temperatures, being so stable that it may be used continuously at temperatures as high as 1550°F. As a result of these characteristics, it finds a wide and varied application. Some of the more outstanding of the varied uses of these steels are that in plates, bars, and castings a large amount is employed by chemical industries for resisting nitric acid corrosion of either dilute or concentrated forms; in strip or sheet form it is applied to such parts as trays or kitchen utensils where severe drawing operations are not necessary; and in strip form, for the molding and trim on automobiles; etc.

The specified analysis for this grade of stainless iron contains 16 to 18 per cent chromium, with carbon from 0.10 per cent to 0.12 per cent maximum, depending upon the application.

These particular alloys are frequently termed *ferritic stainless irons*, meaning that they respond very slightly or not at all to heat-treatment. The explanation is found in the effect of the high chromium, which, in the percentages present, places them outside the gamma loop; hence, they are ferritic and are not hardenable except by cold work. Since the austenite transformation has been suppressed, there will be no phase change on heating or cooling, and recrystallization or grain refinement will not take place. The only effect of heating to high temperatures will be coarsening of the grain structure and, above certain temperatures, this becomes quite drastic.

The effect of high chromium in suppressing the austenite transformation is opposed by the action of carbon, so that with increasing carbon the gamma loop will be extended further to the right and higher percentages of chromium will be necessary to make the alloy ferritic. Thus, with a constant chromium content, an increase in carbon will directly increase the hardening capacity; while, on the other hand, with any given carbon content, an increase in chromium will decrease the hardening.

Heat-treatment.—This grade of steel should be forged between 1950 and 1600°F.; and, should further reduction be necessary, after the metal has cooled at 1600°F., it should again be reheated to 1950°F. Heating is preferably accomplished by a slow preheat to 1400°F. followed by rapid heating to the forging temperature. Since this type of steel is subject to grain growth at elevated temperature, the forging and working temperature should be kept under about 1950°F. It is desirable to keep the grain fine; therefore, the finishing temperatures of the forging or rolling operations are held comparatively low.

Annealing.—The usual temperature range for full annealing of this grade of stainless is 1400 to 1450°F. For full softness, the material should be slowly cooled from this temperature; however, owing to an adverse effect on the impact values by such cooling, the material—depending upon its size and shape—is usually rapidly cooled either by air or water quenching, to avoid this difficulty. This type of treatment cannot be considered a true anneal for reasons previously described, but it does bring about recrystallization of cold-worked areas and transformation of patches of martensite resulting from the small amount of austenite formed at the forging temperature. Partial annealing following hot or cold working may be effected at temperatures between 1250 and 1350°F. Annealing at this lower range produces a fine-grained structure, which is associated with highest strength and ductility.

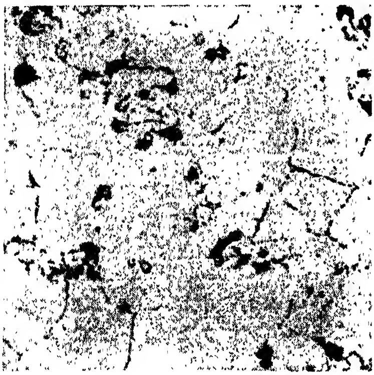

Fig. 10-XXI.—Structure of a 19.10 per cent chromium, 0.11 per cent carbon alloy in the annealed condition. Brinell 179. Etched with aqua regia. × 200. *(Courtesy of Latrobe Electric Steel Company.)*

The annealed structure of the grade is shown in Fig. 10-XXI. It consists of carbide particles imbedded in a matrix of chromium-ferrite crystals.

In general, this material is considered to be nonhardenable by heat-treatment. However, owing to the carbon present, even though in small amounts, there will be a slight reaction from the small amount of austenite formed, so that quenching will bring about an increase in hardness. This hardness, however, rarely exceeds about 230 Brinell, when the carbon is approximately 0.10 per cent, and this is usually insufficient increase over the annealed value to be of commercial value.

Physical Properties.—Table 4-XXI shows the typical physical properties obtained with this material.

Corrosion Resistance.—This type of alloy is resistant to a great many corrosive agents, a list of which is impractically long for this treatise. For further reference see "The Book of Stainless Steels," by Thum. It should be mentioned that this alloy has been found particularly adaptable for nitric acid equipment.

The resistance to high temperatures, which signifies resistance to oxidation, accompanied by the retention of high strength, is quite high in these alloys. A very thin protective scale forms at lower temperatures, which effectively retards further oxidation up to 1550°F. The creep strength at elevated temperatures is, as is shown in Table 4-XXI, in excess of that of ordinary steels.

TABLE 4-XXI.—TYPICAL PHYSICAL PROPERTIES*

Properties	Annealed	Cold-worked (wire)
Ultimate tensile strength, p.s.i.	75,000	100,000–190,000
Yield point, p.s.i.	40,000	
Elongation in 2 in., per cent	27	25–2 (in 10 in.)
Reduction of area, per cent	55	40–20
Impact strength, Izod, ft.-lb.	8–25	
Brinell hardness	175	185–270
Rockwell hardness	85B	90–105B
Stress causing 1 per cent "creep" in 10,000 hr.		
At 1000°F.	....	8500
At 1200°F.	....	2100
At 1350°F.	....	1200
Scaling temperature 1550°F.		

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* WRIGHT, E. C., Rolled Materials, Especially Tubing, "The Book of Stainless Steels," 2d ed., pp. 234–244, E. E. Thum, ed., A.S.M., Cleveland, Ohio, 1935.

Passivating.—During such operations as machining, riveting, forming, polishing, etc., foreign particles may become embedded in the surface of the steel, and, unless they are removed, they will subsequently serve as starting points for local electrochemical attack. To eliminate this possible source of trouble and improve the chemical passivity of the surface of the metal, the finished article should be immersed in warm (120 to 145°F.) 10 to 20 per cent by volume solution of 38° Baumé commercial nitric acid for 30 min. or longer. Following this passivating treatment, the steel should be washed in hot water to remove the acid.

Machining.—Most of the high chromium alloys are difficult to machine. This difficulty has been attributed to two causes: (1) workhardening qualities of the metal, which are inherent in steels of this type; (2) high frictional properties, which are manifested by "galling" and "seizing," and interfere with clean cutting. The first difficulty cannot be avoided, as it is a characteristic of the steel and makes machining difficult, particularly if improperly designed tools are used. The second difficulty may be reduced by decreasing the cutting speed. However, the best method of improving machinability is to modify the composition. Where considerable machining is required and a free-cutting steel is desirable, the composition is modified by adding sulfur and molybdenum or selenium.

Welding.—Seventeen per cent chromium steels can be welded quite readily by electric resistance, electric arc, or oxyacetylene torch methods. This material, however, is subjected to grain growth while at the welding heat, resulting in a coarse-grained structure that cannot be eliminated to any extent by heat-treatment. The coarse-grained welds will thus suffer a corresponding loss in ductility.

Twenty-five Per Cent Chromium Steels.—This grade of steel has its principal applications for service at elevated temperatures and where severe corrosive conditions exist. To be commercially satisfactory for such applications, the chromium content must be in excess of a minimum of 23 per cent. This property is independent of the amount of carbon used. The commercial limits on chromium range from about 20 to 30 per cent. In the case of the wrought products used for their heat and corrosion resistance, the carbon content is generally limited to 0.35 per cent maximum. In the case of castings, the carbon is either held at 0.35 per cent maximum, in applications for resistance to corrosion and high-temperature oxidation, or between 1 and 3 per cent, in applications requiring resistance to high-temperature oxidation and abrasion.

A few of the uses of this material will be mentioned. The lower carbon rolled metal is used for recuperators, heat exchangers, for structural parts in Cottrell precipitators, for bins holding calcined copper ore, etc. The low carbon castings are used for disks, shafts, conveyor chains, and other parts of heat-treating furnaces; for saucer valves in open-hearth regenerators; for annealing boxes and saggers. The medium carbon castings have high resistance, with good wearing qualities. A good example is the rabble blades used in zinc ore-roasting machines.

Other applications are "char" retorts, muffles and muffle linings, parts for domestic coal stokers and oil-burner nozzles, and equipment parts that must work at high temperatures, "carrier sheets" in normalizing furnaces, thermocouple tubes and wells, etc.

The low carbon 20 to 30 per cent chromium steels are ferritic under all conditions of heat-treatment; their structures consist primarily of chromium ferrite plus varying amounts of free iron-chromium carbides. There are no critical transformations, so they are substantially nonhardenable by quenching. These steels are particularly susceptible to grain growth; hence, any heating, particularly heating within the forging range (1900°F. and above), should be as brief as possible to minimize this effect.

The tendency toward large grain size in the cast metal and in the grain growth of the wrought products may be minimized by the use of nitrogen, added in the form of ferrochromium containing nitrogen, in amounts ranging from 0.10 to 0.15 per cent.

The ductility and softness of the low carbon grade in both cast and wrought alloys may be improved by heating to 1550 to 1650°F., followed by rapid cooling in either air or water. Slow cooling should always be avoided, to prevent the so-called *temper brittleness*. This particular condition is developed by slow cooling through the range of about 750 to 1050°F., but is evident only after reaching 210°F., or below. The brittleness can be removed and its recurrence avoided by heating to at least 1100°F., followed by an air cool. The cause for temper brittleness has not been definitely established; however, it would appear that it is the result of hardening by carbide precipitation.

When the carbon content is increased to above 1.00 per cent, the hardness of the steels may be increased by rapid cooling from a temperature of about 1750°F. For increased machinability on this type, the desired condition can be developed by holding for an extended period of time at a temperature of about 1350°F., followed by either furnace or air cooling.

Mechanical Properties.—The properties of cast and wrought chromium steels of this grade are somewhat as follows:[1] An alloy containing 0.14 per cent carbon, 26 per cent chromium, as forged, heated 10 hr. at 1590°F., and water quenched, will have a tensile of 83,000 p.s.i., yield 56,000 p.s.i., elongation 26 per cent, reduction

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[1] MACQUIGG, C. E., 25 to 30% Chromium-Iron Alloys "Book of Stainless Steels," 2d ed., pp. 351–368, E. E. Thum, ed., A.S.M., Cleveland, Ohio, 1935.

of area 61.0 per cent, Brinell hardness number 159, and Izod 2.0. An alloy containing 0.35 per cent carbon, 25 per cent chromium, cast, heated at 1590°F. for 2 hr., cooled to 1100°F., and water quenched, will have a tensile of 58,000 p.s.i., yield 45,000 p.s.i., elongation 2 per cent, reduction of area 3.0 per cent, Brinell hardness number 149, and Izod 2.0; and an alloy containing 2.00 per cent carbon, 22 per cent chromium, as forged, heated to 1470°F., slowly cooled to 1100°F., then air cooled, will have a tensile of 110,000 p.s.i., yield of 100,000 p.s.i., elongation 9.0 per cent, and 12.5 per cent reduction of area.

Corrosion.—One of the most important outstanding properties of these alloys of 26 to 27 per cent chromium is their resistance to oxidizing atmospheres up to temperatures 2100°F. (1150°C.) or higher. They maintain their surfaces indefinitely within this temperature limit, apparently by accumulating a thin, tough, adherent scale, which appears to act as a protection against further surface deterioration. This scale is so adherent that it will not flake off even when subjected to drastic temperature changes. The alloys are perfectly safe in oxidizing atmospheres up to the melting points but in the presence of reducing gases, such as carbon monoxide and flue gases, above about 2200°F. (1200°C.), they are rapidly deteriorated.

One point of particular importance is in regard to high chromium steels and their resistance to oxidizing conditions, which holds for both liquids and gases at either room temperature or elevated temperatures. If the attacking medium is not capable of developing this protective oxide coating, the medium will generally prove destructive in its action. As a precautionary measure, unless the conditions that are to be imposed on the metal are known without the possibility of error in judgment regarding its behavior, actual tests should be made to cover the given conditions.

18-8 Chromium-nickel Steels.—Of the corrosion-resisting steels containing both chromium and nickel, the so-called 18-8 type is by far the most widely applied. The reason for its wide application is due to its properties of high corrosion resistance, resistance to scaling at high temperatures, strength at high temperatures, and formability both hot and cold.

The typical composition of this grade of steel will contain carbon either 0.08 to 0.20 per cent, or 0.08 per cent maximum, chromium 17.00 to 19.00 per cent, and nickel 7.00 to 9.00 per cent, 1.25 per cent maximum manganese with the higher carbon range and 2.00 per cent maximum with the lower. It appears that for best results the chrome and nickel should total 26 to 27 per cent.

The two ranges in carbon content are made to meet varied specifications. The low carbon material is less susceptible to the so-called *intergranular corrosion* when exposed for long periods of time at high temperatures and is somewhat superior from a corrosion-resistant viewpoint.

It appears that, in order to obtain the most desirable mechanical properties, the carbon content should range from 0.08 to 0.12 per cent. In the so-called *dead-soft* or fully annealed condition, for a constant testing speed, it is found that as the carbon increases, the tensile and yield strengths will be increased, with the elongation not materially affected within the range of 0.08 to 0.15 per cent carbon. The yield strength will be more affected than any of the other properties. Of all the elements present in this type of steel, carbon is the most important in the control of physical properties.

Variations in the other elements on the physical properties, all other elements remaining constant with respect to an analysis of about 0.10 carbon, 8.0 nickel, 18.0 chromium, 0.50 manganese, in the "dead-soft" condition, with constant testing speed, are as follows:

1. An increase in nickel above 6.50 per cent will cause the tensile strength to be reduced, the yield to be reduced but not appreciably, and the elongation to drop.
2. Increase in chromium above 16.5 per cent will have no appreciable effect on the physical properties.
3. Increase in manganese from 0.50 to 1.50 per cent will have approximately the same effect as nickel.
4. Increase in the nitrogen content from about 0.05 to 0.15 per cent will bring about the same effect as carbon.

The physical properties of the aforementioned balanced analysis will be approximately as follows:

Yield strength	42,000 p.s.i.
Tensile strength	93,000 p.s.i.
Elongation in 2 in	58 per cent

Various manufacturing or service conditions have brought about the development of modifications of the 18-8 analysis.

Molybdenum is added in the amount of 2 to 4 per cent, with nickel as high as 14 per cent, to aid in fabrication and to prevent damage from local, isolated, so-called *pinhole* corrosion, which develops when the usual analysis is exposed to sea water, some chlorides, and vapors of acetic acid. Silicon, when added in amounts ranging from 1 to 3 per cent, increases the general oxidation resistance. Additions of copper (up to 2 per cent) and copper and molybdenum (about 1 per cent each) are used to improve the corrosion resistance toward certain media. The additions of sulfur (0.20 to 0.40 per cent), selenium (0.20 to 0.30 per cent), or lead (0.12 to 0.25 per cent) may be used to improve the machinability.

Heat-treatment. Annealing.—From the previous study of equilibrium diagrams involving the effect of chromium and nickel on the transformation, we know that this alloy is not hardenable by heat-treatment and that it is, except after cold workings, as will be subsequently noted, fully austenitic, and nonmagnetic. The austenite is unstable, but persists because of sluggishness toward any change. In many cases this instability is unimportant, but when the application involves heating to temperatures between about 750 to 1500°F. or cold working, some decomposition of the austenite takes place and the degree of instability becomes an important factor.

Annealing in this type of alloy, to produce maximum softness, involves heating to temperatures of from 1800 to 2100°F., holding for the proper length of time (depending upon the temperature), followed by rapid cooling. The time at temperature usually involves a matter of several minutes at the higher temperature and longer at the lower, to produce the desired result. This treatment will produce a completely austenitic structure with the carbides in solution. The higher temperatures produce a slight increase in ductility with a slight decrease in tensile strength; but, at temperatures beyond 2000°F., grain growth is so rapid that unless precaution is taken the economy gained through shorter heating time may be lost through adverse properties produced.

Figure 11-XXI-*A*, *B*, and *C* shows typical microstructures of an 0.08 per cent carbon 18-8 alloy in the hot-rolled and the heat-treated condition, respectively. The structures are all austenitic; the polyhedral crystals in the heat showing pronounced twinned-

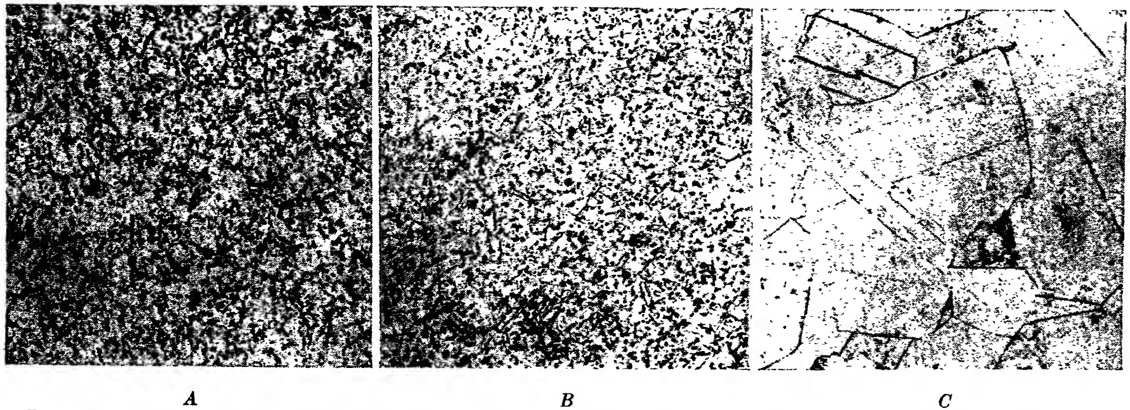

FIG. 11-XXI.—Structures of an 18.70 per cent chromium, 7.96 per cent nickel, 0.08 per cent carbon alloy. *A* is a hot-rolled structure, Brinell 217; *B* has been water-quenched from 1750°F., Brinell 163; and *C* has been water-quenched from 2100°F., Brinell 143. All etched with aqua regia × 200.

grain characteristics. The effect of heating temperature on the grain size developed at the low and high ends of the range is apparent.

Cold Work.—The effect of cold work on the mechanical properties is shown in Table 5-XXI. As is evident, the cold work increases the strength and hardness values very greatly, while reducing the ductility and elongation. The peak values of 300,000 p.s.i. tensile, 250,000 p.s.i. yield, are properties attainable with the severest cold work.

The degree of reduction will, of course, have a direct effect on the strength figures obtained; however, the amount of cold reduction necessary to attain a given set of physical properties is found[1] to vary according to the sum total of the elements nickel, chromium, and carbon in the alloy, and their relation to one another. Thus, the mechanical properties may be adjusted by the proper choice of a given analysis. This relationship may be explained by considering several facts. The austenitic structure observed at room temperature is found to be unstable and tends to transform slowly to the ferritic condition on the appli-

TABLE 5-XXI

Properties	Annealed water-quenched from 2000°F.	Cold-worked (wire)
Ultimate strength, p.s.i.	80,000–95,000	105,000–300,000
Yield point, p.s.i.	35,000–45,000	60,000–250,000
Elongation, per cent in 2 in.	55–60	50–2 (in 10″)
Reduction of area, per cent	60–75	65–30
Impact, ft.-lb. Izod	75–110	
Brinell hardness	140–180	170–460
Rockwell hardness	B80–84	C5–47
Stress causing 1 per cent creep in 10,000 hr.		
At 1000°F.	....	17,000
At 1200°F.	....	7,000
At 1350°F.	....	3,000
At 1500°F.	....	850
Scaling temperature 1650°F.		

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[1] KRIVOBOK and LINCOLN, Austenitic Stainless Alloys, *Trans. A.S.M.*, **25**, 637–689 (1937).

cation of varying amounts of cold work. The transformation is evidenced by the fact that the alloy gradually increases in its response to a magnet (ferrite being magnetic). Since the degree of instability is a function of the composition, various mechanical properties can, therefore, be secured by selection of analysis. The effect of a decrease in either nickel or carbon, at lower amounts of cold deformation, is to favor the formation of ferrite, while an increase in nickel makes the formation of ferrite more difficult. (For certain applications this is advantageous —18-12 alloys. With the high nickel percentage, the austenite-to-ferrite transformation is almost entirely suppressed and the rate of hardening on cold working is correspondingly decreased.)

The cold-worked structure of the 18-8 austenitic alloy is shown in Fig. 12-XXI. The deformation has proceeded, as in all metals, by the mechanism of slip along crystal planes.

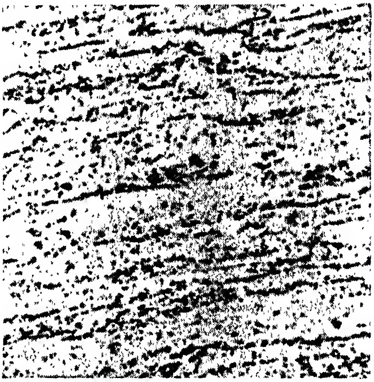

FIG. 12-XXI.—Structure of an 18.70 per cent chromium, 7.96 per cent nickel, 0.08 per cent carbon alloy, after cold working. Brinell 197. Etched in aqua regia × 200. (*Courtesy of Latrobe Electric Steel Company.*)

After cold work, heating to a temperature of 1800°F. or higher will bring about complete removal of stresses, recrystallizaton, and full softening, as well as carbide solution. Cooling from the annealing temperature should be rapid in order to avoid carbide precipitation in the 1500 to 750°F. range. A reheat up to 750°F. after cold working is found to improve the yield, ultimate strengths, and the ductility (stress relief), with no appreciable softening.

Stabilized 18-8.—Mention has been made of the fact that after exposure to moderately high temperatures of about 750 to 1500°F. there is a precipitation of carbides. This precipitation takes place in the grain boundaries and causes one of the few shortcomings of the austenitic 18-8 steels (this includes the higher alloys, such as 18-12, 25-12 and 25-20), that of an intergranular corrosion in environments that ordinarily would have no effect upon the normal alloy. This corrosive effect is sometimes noted near a weld. Since welding does affect a thermal gradiant in the base metal adjacent to the weld, it will cause an accompanying carbide precipitation in that portion which has been heated in the sensitive-temperature range. Such precipitation frequently causes the metal near the weld literally to disintegrate in use with solutions that will not attack the remainder of the metal. Whenever the metal was found to be in this sensitive condition to intergranular attack, it could be restored to its normal corrosion-resisting state by heating to a higher temperature, *i.e.*, 1650 to 2000°F., followed by rapid cooling.

Characteristics of Sensitive Material.—It must be noted that these characteristics refer not to the normal or general corrosion resistance of stainless steel but exclusively to the intergranular corrosion in a medium that causes no loss of weight on the normal metal but does cause severe intergranular attack upon sensitive material. Investigation has brought out the following factors:

The carbon content of the alloy has considerable importance. The carbon that has been dissolved at the high temperatures and is retained in a supersaturated solution at ordinary temperatures is rejected to the grain boundaries in the form of a chromium-rich carbide when heated at intermediate temperatures. The carbide, on forming, will draw upon the chromium content of the metal adjacent to a carbide particle and will, therefore, take away the chromium until it is below the amount necessary for normal corrosion resistance. Therefore, a regular pathway is opened up through the metal along the grain boundaries wherever there has been carbide precipitation.

It should be emphasized that this phenomenon is dependent upon the peculiar fact that the relief of carbon supersaturation in the austenitic stainless alloys occurs by rejection to the grain boundaries and to the slip planes. There is an apparent lack of nuclei for carbon precipitation in the body of the grains.

The composition is found to have a decided influence upon the rate of attack. It is found that both the low carbon as well as the high carbon 18-8 steels are attacked, but apparently in the case of the higher carbon steels a much shorter time is required. If the time of heating is long, the carbon content has much less influence than when the time is short. The severity of the sensitization is approximately proportional to the carbon content of the alloy, with little regard to the chromium and nickel within the normal variation.

The effect of cold work upon the sensitization is important. It is found that cold-worked metal becomes sensitized much more rapidly than annealed material, but the sensitization is never so severe in the case of the cold-worked metal. It also has a very rapid recovery of resistance to intergranular penetration by continued heating at the temperature that induced sensitization. From a study of the structure it is found that, on heating, the carbides are precipitated on all the slip planes as readily as they are upon the original grain boundaries. Thus, the carbon is quickly exhausted by the general precipitation throughout the grain and there will be no point at which the chromium has been greatly reduced. Chromium diffusion, although slow, has only a short distance to travel; hence, the rapid recovery of the cold-worked metal to corrosive attack. Cold working, therefore, induces the regular age-hardening precipitation phenomena in a

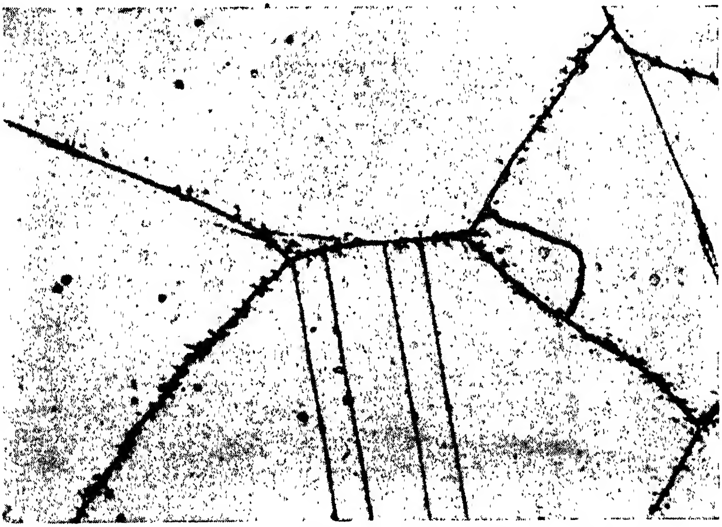

FIG. 13-XXI.[1]—Photomicrograph of carbide particles precipitated in grain boundaries. Taken at × 3000. (*Courtesy of Research Laboratory, United States Steel Corp.*)

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[1] BAIN, ABORN, and RUTHERFORD, The Nature and Prevention of Intergranular Corrosion in Austenitic Stainless Steels, *Trans. A.S.S.T.*, **21**, 481–509 (1933).

material that otherwise fails to provide precipitation nuclei except at the grain boundaries.

Figure 13-XXI. shows the carbide particles that have been precipitated in the grain boundaries by reheating in the sensitive-temperature range. The matrix is merely a uniform solid solution.

Prevention of Intergranular Corrosion.—From a review of the factors affecting the corrosion resistance of this metal it is quite apparent that the first means of prevention would be to reduce the carbon percentage below the solubility limit at low temperatures. It is difficult to determine this from the constitutional diagram, but it is found to be approximately 0.02 per cent. At the present time, however, such a procedure is regarded as impractical. Another method is to bring about a general carbide dispersion by cold work, followed by a carbide precipitation treatment at a temperature between 1250 and 1400°F.

Probably the most widely used method of preventing intergranular susceptibility, at the present time, is to incorporate in the alloy a strong carbide-forming element, such as titanium, columbium, or uranium, with a suitable heat-treatment. Alloys to which such additions are made, columbium and titanium being the most used, are termed *stabilized.* The evidence seems to indicate that these elements are very much more active carbide formers than is chromium, so that during the stabilizing treatment carbides of these elements are formed in preference to chromium. Thus the chromium is left free to remain in solid solution, where it can effect full corrosion resistance.

It might be expected that a prolonged heating of, for example, a titanium-bearing 18-8 at, say, 1000°F. would result in establishing equilibrium for that temperature through the precipitation of titanium carbide, but in 18-8 alloys containing the proper amount of titanium a small amount of chromium carbide also is formed. Under these conditions a sensitization may actually be detected. Stabilization of the titanium-bearing 18-8 alloys, fortunately, can be brought about by a practical treatment of heating from 2 to 4 hr. at approximately 1560°F. (850°C). When maximum softness is desired, a high-temperature quench may be used prior to the stabilizing treatment.

Corrosion Resistance.—When we speak of corrosion resistance it must be noted that there is no single metal and no alloy developed that remains unattacked in all solutions. In general, only the rarer metals, such as gold and platinum, offer any considerable resistance to corrosive attack. The chromium-iron alloys and those with nickel are quite obviously the cheapest of our alloys that offer ample resistance to the more common and most active corrosive conditions.

A complete list of the industrial agents that do not attack a clean surface of 18-8 would be unduly long, but the accompanying list will give a general impression of the kinds of environment in which it is highly corrosion-resistant (U.S.S. Stainless and Heat Resisting Alloys):

Environment in Which 18-8 Is Corrosion Resistant

- Acetic acid, cold, any concentration
- Acetic acid, hot, up to 10 % approx.
- Alkaline solutions, in general
- Bichloride of mercury, dilute (usual antiseptic strength)
- Carbolic acid
- Carbonated water
- Citric acid, cold, moderate strength
- Copper sulfate
- Fruit and vegetable juices
- Hydrogen peroxide
- Zinc chloride, cold
- Hydrogen sulfide
- Laundry solutions, with few exceptions
- Milk and dairy products
- Nitric acid
- Photographic solutions
- Salt solution
- Sea water
- Sulfuric acid, cold, very slight action
- Sulfurous acid
- Wood pulp
- Yeast
- Zinc sulfate

Through the formation of the inert surface condition, 18-8 is resistant, at ordinary temperatures, to the atmosphere, with its usual traces of corrosive gases and moisture, to chemical solutions of a large variety; and it has high creep strength, along with high resistance to oxidation, at elevated temperatures. This latter condition is due, as before described, to the thin adherent scale, which, when once formed, oxidizes very slowly. 18-8 is found to scale comparatively little below 1600 to 1650°F.

Higher Alloys of Chromium-nickel-iron.—The alloys of this particular group are particularly important in the heat- and corrosion-resistant fields of service. For such service the chromium is carried quite high, 18 to 30 per cent, in order to ensure adequate high temperature and corrosion resistance, while the nickel with this range of chromium is held to a minimum of 8 per cent, which is the lowest amount that will make the austenite reasonably stable, and may run as high as 65 per cent. The carbon content is usually kept below 0.20 per cent in the wrought alloys, and between 0.40 and 0.70 per cent in the cast. The alloys containing a higher nickel than chromium content are not so susceptible to the brittleness effect caused by carbide precipitation, but excessive carbon does develop a serious loss in high-temperature strength. The alloys of higher chromium than nickel, however, are susceptible to brittleness through carbide precipitation with relatively high carbon percentages; they are also susceptible to brittleness through carburization, the rate increasing with decreasing amounts of nickel. All these alloys have much better high-temperature strength than the chromium steels of equal chromium content previously discussed.

Some of the usual analyses of this group of alloys are: 25 per cent chromium, 12 per cent nickel; 28 per cent chromium, 10 per cent nickel; 25 per cent chromium, 20 per cent nickel; 15 per cent chromium, 65 per cent nickel; 15 per cent chromium, 35 per cent nickel; 20 per cent chromium, 25 per cent nickel; the balance in all cases is principally iron.

Since all the alloys in this group are austenitic and, therefore, not heat-treatable, they will not be discussed further.

WEAR-RESISTING STEELS

Austenitic Manganese Steels.—The austenitic, or Hadfield's, manganese steel is probably best known in its application where the steel must resist very heavy wear. In general, wherever hard materials are to be dug, crushed, handled, or moved, manganese-steel parts are widely used to resist the wear encountered in these operations. No other steel has been found that will equal it in a great number of such applications.

The commercial austenitic manganese steels are within the following limits:

Carbon	Manganese	Silicon	Sulfur	Phosphorus
1.00–1.40	10.00–14.00	0.30–1.00	0.05 Max.	0.10 Max.

For most reliable results the lower limit of manganese is usually set at 11.00 per cent instead of 10.00 per cent, as the lower limit steels are somewhat lacking in toughness and strength.

Heat-treatment.—In the cast condition these steels are found to be quite brittle. In other words, when they are slowly cooled from the casting temperature, considerable cementite, as manganitic cementite, separates from the austenite and forms a network around the grains. This cementite is extremely hard and rather brittle. Annealing of these steels is, therefore, obviously illogical. To strengthen and toughen the metal, the pieces are heated to 1830 to 1940°F. and then water quenched. This treatment serves to take the carbide segregations in solution in the austenite. By rapid quenching they are more or less prevented from reprecipitating. When the composition is within the correct limits, this treatment will produce a uniform austenitic structure (Fig. 14-XXI) and a high degree of strength and toughness. After this treatment, the tensile strength will range from 135,000 to 145,000 p.s.i., with an elongation of 50 to 60 per cent. By varying the factors of temperature, duration of heating, and rate of cooling, it is possible to obtain a wide range of physical properties. The static strength and ductility are governed to a large degree by the amount of original free carbide that is taken

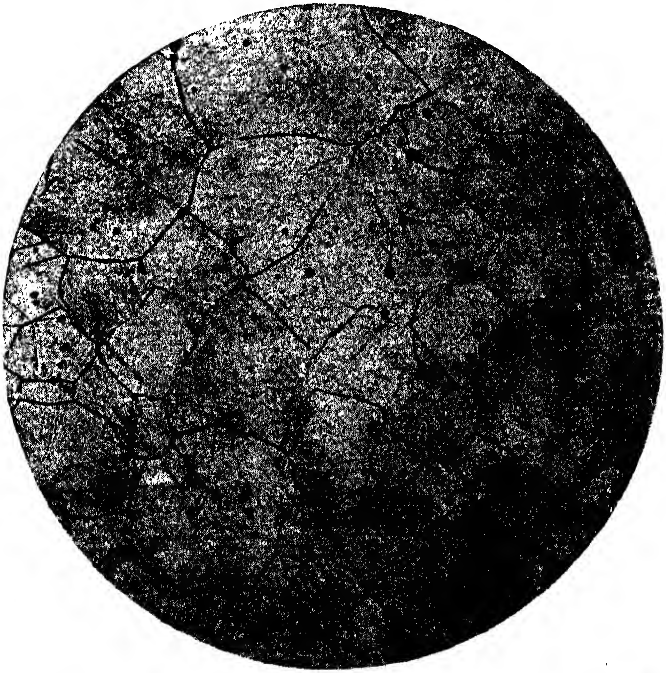

FIG. 14-XXI.—Structure of a ¾ in. round, rolled 12 per cent manganese steel, heated to 1940°F. (1060°C.) and quenched in water. Taken at × 265. Aqua regia etched. (*Courtesy of J. H. Hall, American Institute of Mining and Metallurgical Engineers.*)

into solution and retained there by water quenching. The original as-cast structure and the physical properties—which means essentially the condition of the carbides—are greatly modified by the method of casting, the size of the casting, and the temperature at which the steel has been poured. If the pouring temperature has been "hot," the steel produced will have a coarse crystalline structure, which is not changed by subsequent heat-treatment. Figure 15-XXI illustrates the coarse structure of the steel that has been poured hot, heat-treated, and cold-worked. The lines extending through the grains are the familiar slip lines of cold-worked material, austenite in this case. Hot working will improve the properties somewhat, but the most useful properties are obtained by heat-treatment.

Fig. 15-XXI.—Structure of a heat-treated, hot cast manganese steel rail. × 60. Etched with nitric acid and alcohol. (*Courtesy of J. H. Hall, American Institute of Mining and Metallurgical Engineers.*)

If the heat-treated steels were to be reheated for a long time at 660°F. or for a shorter time at a higher temperature, they would become very brittle. Thus, it is impossible to temper these steels for the relief of the stresses that are set up during the cooling. It becomes important, in view of this fact, that there should be reasonable uniformity of sections between different parts of a casting so as to avoid severe warping and even breaking of the casting in treatment. The same conditions will hold true in regard to the thickness of section that may successfully be heat-treated. Sections of about 6 in. or more, owing to the unequal cooling rates of the inside and outside, set up stresses that are so severe as to cause cracking.

One of the most interesting and important properties of austenitic manganese steel is its ability to be hardened by a slight amount of cold work. Several examples will serve to show the effects of cold working on the hardening power and wear of this steel.

In the as-cast condition, the Brinell hardness of the steel is about 185, while in the heat-treated condition it is between 180 and 200. This, as can be seen, is only moderately high. The correctness of this figure is shown by the fact that the steel can easily be dented by a hammer or nicked by a chisel up to a certain point. The initial blows of the hammer produce a considerable dent but, as the blows are continued, the impression becomes smaller and smaller until a point is reached where the blows hardly mar the steel. This is due to the effect of the cold working, which increases the hardness. A Brinell test shows that the hardness has increased to 450–550. Cold working of all kinds whether it be hammering, pulling, bending, squeezing, or twisting is probably responsible for this property, *i.e.*, the ability to endure severe abrasion coupled with heavy pressure or blows.

As an example of wear and abrasion, let us consider a jaw crusher where both effects are present. In this application, tests have shown that manganese steels will outlast, from five to ten times, metals whose initial hardness is much greater. The initial rate of wear of this steel is found to be somewhat greater than the wear after a few tons of material have been crushed. This is due to the hardening effect brought about by the kneading action of the material being crushed.

In wear of different kinds, where cold working is largely absent so that there is no surface hardening of the metal, these steels will not outlast other metals. Thus, in the application of austenitic manganese steels we must consider whether there is sufficient pounding or pressure to produce a useful degree of hardening.

Applications.—Typical applications of cast high manganese steel are railway and trolley frogs, crossings and switches, rock-crusher parts, steam-shovel dippers and teeth, dredge buckets and lips, conveyor chains, sprockets, gears and pinions of many types, ball- and tube-mill liners, tractor shoes, and many others.

Other Wear Steels.—The casehardening steels of both the simple carbon or alloy grades have been discussed in a previous section, so they need not be considered again.

A rather important group of steels includes those used by the ball- and roller-bearing industry in the manufacture of races, rollers, and balls. These steels must be of a high degree of cleanliness and, therefore, require carefully controlled practice. Bar stock for the automatic machining of races and for the cold heading of balls is usually specified in the spheroidized, annealed structure. For forgings, the bars are specified in the as-rolled condition.

The requisites of these steels generally demand high hardness, load-carrying ability, and wear resistance. The steel most commonly used for races is the S.A.E. 52100. The balls are also made from high chromium, high carbon steels, the carbon similar to the S.A.E. 52100, but the chromium percentage, lower.

Where shock is encountered, which demands a tough interior and hard surface, carburizing steels, such as S.A.E. 4615, 5120, 3115, 3312 and 6120, may be employed.

TABLE 6-XXI.—TYPICAL COMPOSITION

Elements	Bearing steel 52100	Ball steel
Carbon	0.95–1.10	0.90–1.10
Manganese	0.20–0.50	0.50 (max.)
Phosphorus (maximum)	0.03	0.025 (max.)
Sulfur (maximum)	0.035	0.025 (max.)
Silicon	0.15–0.30	0.15–0.30
Chromium	1.20–1.50	0.75–1.20

Heat-treatment.—The hardening treatment consists in heating —for large sections, up to 1550°F.; for small sections, 1450°F.— followed by quenching in light oil. Tempering is carried out at a low temperature of around 300°F. Properly treated steels will show a Rockwell C of 64 to 66. Ordinary Brinell readings will give fictitious values, owing to the fact that the Brinell ball is made of this type of steel.

TOOL STEELS

In the study of tool steels, let us first consider briefly the general principles that these materials should embody.

Hardness, of course, is an essential of every cutting tool. This hardness must be materially greater than that of the material that is being cut, but there is usually a limit beyond which a further increase in hardness will be of little use. For example, a steel saw for cutting wood requires only moderate hardness and it should be able to be sharpened with a file. Any increase in hardness above this point would add little to the wear resistance of the saw and would make it more difficult to sharpen.

Strength also plays an important part in cutting quality, in that if one material is to cut another material at commercial speeds it must have a strength greater than that of the material being cut. Toughness is required in a degree just sufficient to enable the tool to resist the stresses and shocks encountered in its manufacture and its use, without breaking. An additional amount of toughness is required to ensure that the cutting edge will not show a prohibitive amount of cracking, chipping, and crumbling.

Workability, or the ease with which the material may be fabricated into tools, also has an effect on the degree to which a particular material is used. For example, an intricate punch-press die could hardly be fabricated from some of the hard alloys.

Cost is always an important factor, in considering which we must take into account, among other things, the life of the tool and the job on hand.

Material for cutting tools may, in general, be divided into two classes: (1) steels, and (2) nonferrous materials, including such substances as diamonds, most abrasives, the hard carbides, and some of the hard metal alloys.

Classification.—The tool-steel situation is so complicated that it becomes necessary to lay out a simple system of classification.[1] The following classification is developed on the basis of wear

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[1] This classification and much of the following data have been taken from the bulletin, Carpenter Matched Tool Steels, published by the Carpenter Steel Company, Reading, Pa.

resistance, toughness, freedom from size change, hardness, and red-hardness. Such a classification is intended to bring out the relative advantages of the various types of tool steel and the reasons for their use on the different types of jobs. It should be emphasized, however, that, in addition to this particular grouping, other special steels are available to meet certain specific service conditions that will not be covered by the steels herein studied.

Our basis for this classification is a special 1.10 per cent carbon, water-hardening tool steel. With this steel as a key, let us draw a diagram (Fig. 16-XXI) in which the steels in the direction of the arrows are required to have the properties as indicated. To the north, the steel has greater wear resistance; to the south, greater toughness; to the west, safe hardening and the ability of retaining its size and shape more accurately during the quench; and to the east, red-hardness, or the ability to retain its hardness at high temperatures.

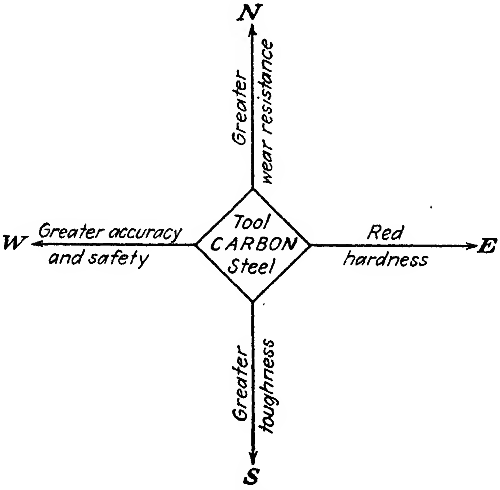

Fig. 16-XXI.

Frequently the toolmaker desires two things, *i.e.*, he may want greater wear resistance and greater hardening accuracy, as well. To meet these requirements, it is necessary to go north on the diagram, for wear resistance, and then west, for accuracy. Similarly, the steels required may have any other combination of these four factors: greater toughness and hardening accuracy;

greater wear resistance and maximum hardness; and red-hardness with maximum toughness. According to this diagrammatic scheme, only nine steels are required to cover any tool problem that might be encountered in service. These nine steels will now be studied in detail.

Carbon Tool Steel.—The carbon tool steel or water-hardening steel is probably the oldest and simplest form. In later years its once almost universal field was considerably limited until, with the more recent developments in methods for controlling hardness-penetration and grain-size properties, its usefulness has again been greatly widened. Probably it is now more generally used than any other type of tool steel. It is especially well adapted to those classes of service where a hard surface is required to resist wear, as well as a tough interior to resist shock or fracture. Carbon plays a large part in determining the properties of all grades of steel, but in straight carbon tool steel it is the dominant factor. The higher the carbon, the more care must be taken in the forging, hardening, and tempering.

A typical analysis for the key steel in our diagram is as follows:

Elements	Per Cent
Carbon	1.10
Manganese	0.20
Silicon	0.20
Alloys	0.00

Heat-treatment. Annealing.—This grade of steel may be annealed by heating slowly to a temperature of 1350 to 1375°F., holding at that temperature until all parts are thoroughly and uniformly heated, and then cooling slowly in the furnace. Such annealing will produce a Brinell hardness of 160 to 190.

Hardening.—To be hardened, this steel should be slowly and uniformly heated in a neutral or slightly oxidizing atmosphere to 1450°F. and quenched in 5 to 10 per cent salt brine. For the three water-hardening steels to be discussed, brine is much better than fresh water.

In such parts as coining dies, striking dies, heading dies, etc., which are under heavy pressure, the hardening temperature can be raised to 1500 or even 1600°F. to increase the penetration of hardening to prevent sinking.

Drawing.—The effect of drawing temperatures on this steel after brine quenching from 1450°F. is shown in Table 7-XXI.

Higher draw temperatures are not shown, as this steel should not be drawn softer than about C-62 Rockwell. If the resultant toughness is not sufficient, it becomes necessary to use a steel with greater toughness.

TABLE 7-XXI

Drawing Temperature 1 Hr. at Heat	Rockwell Hardness
Not drawn	C 66–67
200°F	66–67
300°F	64–65
350°F	63–64

The size and shape of the tool will govern the length of time at the draw temperature. For securing maximum toughness, Table 8-XXI gives the best drawing temperature for all combinations of hardening temperature and drawing time.

TABLE 8-XXI

Hardening temperature	Drawing time				
	15 min.	1 hr.	4 hr.	8 hr.	24 hr.
1450°F.	375°F.	350°F.	325°F.	300°F.	275°F.
1500°F.	400°F.	375°F.	350°F.	325°F.	300°F.
1550°F.	425°F.	400°F.	375°F.	350°F.	325°F.
1600°F.	450°F.	425°F.	400°F.	375°F.	350°F.

For example, if the tool is hardened at 1500°F. and it is decided to draw 4 hr., draw at 350°F., etc. After drawing for the proper time, let the tool cool in air.

Characteristics and Uses.—Some of the most valuable characteristics of this steel are that it will do more jobs than any other steel; it is easier to machine; hardness can be imparted to it that is greater than that of many of its alloy competitors; it has good toughness, especially when subjected to shock; it is easily fabricated; its hardness or temper may be varied quite easily; and, it has low cost. These qualities have made it especially applicable for hand tools, tools subject to severe shock, and tools that encounter only low operating temperatures. Some of its applications are cold-working dies used for blanking, stamping, heading, trimming, embossing, engraving, coining, etc. It is also well adapted to such edge tools as punches, cutters, shear blades, chisels, hand reamers, bitstock drills, small wire-size drills, hand taps, thread-cutting dies, and numerous kinds of tools for easily machined metals and alloys.

Water-hardening Steel (Water-tough).—To meet the requirements of toughness with high strength, south on the diagram, a steel containing manganese, silicon, and molybdenum of the following analysis is suggested.

Elements	Per Cent
Carbon	0.50
Manganese	0.40
Silicon	1.00
Molybdenum	0.50

This steel is the strongest and toughest tool steel known. A tensile test on a standard specimen, after proper treatment, will show

Breaking strength	323,000 p.s.i.
Elongation in 2 in	4.5 per cent

Even with this great strength, this steel is ductile enough to stretch more than 4 per cent before breaking. At C-59 Rockwell, this steel will bend before it will break.

This steel is found to acquire a hard exterior and a tough center when hardened in medium and large-sized pieces (sizes under ¾ in. will harden clear through); to expand more in hardening than will the straight carbon grade, owing to the fact that it hardens deeper (this steel would, therefore, not be used for tools that must hold their shape and size accurately in hardening); and to decarburize more than any other steel that we shall study, because of its silicon and molybdenum content. This can be avoided if proper heating and furnace control are used; or, if the tools are ground after hardening, decarburization does not matter.

Heat-treatment. Annealing.—For annealing, this grade of steel should be heated quite slowly and uniformly to 1350 to 1400°F. in a container with clean cast-iron borings or other packing material, so as to hold scaling and decarburization to a minimum. In the fully annealed condition, this steel will show an average Brinell hardness of 190 to 220.

Hardening.—To harden this steel, heat it uniformly to 1550°F. and quench it in 5 to 10 per cent brine. For small parts or parts of intricate shape, oil quenching is recommended. The furnace atmosphere should be carefully controlled to prevent surface decarburization. Heat in a lead pot when practicable or, if a furnace is used, adjust it to a slightly reducing atmosphere.

Drawing.—After water-hardening (brine) from 1550°F., the hardness values in Table 9-XXI will be secured in drawing.

TABLE 9-XXI

Drawing Temperature 1 Hr. at Heat, °F.	Rockwell Hardness
Not drawn	C 61–63
200	61–62
250	60–61
300	59–60

It is recommended to draw all water-quenched cutting tools to 275 to 300°F. for 1 hr. or longer, according to the size. Shanks of pneumatic chisels are drawn at 900°F.

Uses.—This steel should be used when the straight carbon steel is not tough enough. As a general rule, do not draw the carbon steel below C-62 Rockwell but change to a steel of the water-tough type.

Among the common applications of this grade are chisels, rivet busters, and boilermakers' tools, cold sets, swages, heavy-duty punches, pipe-cutter wheels, driftpins, shear blades, nail sets, and machine parts subject to shock.

All these applications demand a combination of hardness and toughness that can be obtained with a carbon steel only by lowering the carbon percentage and drawing back too far. The water-tough steel is far the best in these cases.

Water-hardening Steel (Water-wear).—On the wear-resisting side of our diagram, north on the diagram, there is a water-hardening tool steel that will meet requirements of wear just as the manganese, silicon-molybdenum satisfied the requirements on the tough side. This steel has the following typical analysis:

Elements	Per Cent
Carbon	1.30
Manganese	0.30
Silicon	0.40
Tungsten	3.50

The carbon and tungsten of this steel combine to form complex high tungsten carbides that are extremely hard. These carbides, in properly treated steel, are found to be uniformly distributed throughout the matrix and, in effect, take the brunt of the wear. This steel is found to resist wear and abrasion from four to ten times longer than will plain carbon tool steels. Aside from great wear resistance, this steel behaves much like a straight carbon tool steel; but after heat-treatment it acquires a peculiar slippery-hard surface, which is so difficult to scratch that an ordinary emery wheel can scarcely cut it.

This steel is a shallow-hardening one and hardens with a somewhat thinner case than does the straight carbon. Owing to this property and to its being a water-quenched steel, this steel tends to shrink in hardening. This frequently is a useful property in draw dies, which can be shrunk enough in rehardening to take up the wear and thus allow them to be reused for the same size. Obviously, this steel should not be used for tools that must hold their size and shape accurately in hardening.

Heat-treatment. Annealing.—To anneal this steel, pack it in a suitable container with packing material and heat uniformly to 1440 to 1460°F., then cool it slowly in the furnace. Properly annealed steel of this grade will show an average Brinell hardness of 200 to 230.

Hardening.—To harden, heat uniformly to 1525 to 1600°F. and quench in 5 to 10 per cent salt brine. The hardening temperature to be used varies with the section size and the hardness penetration desired. Since this steel is shallow-hardening, it should never be quenched in fresh water except in cases where a "flush" is used; for example, for hardening a cavity in a die or for flushing large rings. This precaution is necessary so as to avoid soft spots.

Drawing.—The average Rockwell hardness value to be secured on the steel, following a brine quench from 1550°F., is shown in Table 10-XXI.

Table 10-XXI

Drawing Temperature 1 Hr. at Heat, °F.	Rockwell Hardness
Not drawn	C 67–68
200	66–67
300	66–67
350	65–66
400	64–65

The best drawing temperature for maximum hardness and toughness is 350 to 375°F. If greater toughness is needed, it would become necessary to revert to the straight carbon steel.

Uses.—We have now been able to place in our diagram the steel for all water-hardening tool-steel jobs where the requirements demand greater wear resistance and greater production than the straight carbon steel will deliver. This steel has been successfully used in the following applications: draw dies, shaving dies, cold-extrusion dies and punches, mandrels, forming tools, burnishing tools, cutting tools for brass and nonferrous metals, finishing reamers, spinning tools, etc.

Manganese Oil-hardening Steel (Oil-hard).—In this steel we are going west on the diagram; so that the requirements are greater accuracy and safety in hardening. Such demands can be met only with an oil-hardening steel for the following reasons:

1. Because it holds size in hardening.
2. Because it warps less in hardening.
3. Because it is safer to harden in intricate shapes.

The typical analysis of the oil-hardening tool steel that possesses maximum safety from hardening cracks when quenched in intricate sections and will harden with a minimum change of size and warpage is as follows:

Elements	Per Cent
Carbon	0.90
Manganese	1.60
Silicon	0.25

Heat-treatment. Annealing.—To anneal this steel, pack it in a suitable container with packing material and heat uniformly to 1375 to 1400°F., then cool slowly in the furnace. Properly annealed steel of this grade will have an average Brinell hardness of 170 to 200.

Hardening.—The critical point for this steel is below 1350°F., so that we should be able to harden at any temperature higher than this, but an oil-hardening steel of this type will hold its size during hardening only if hardened at one temperature. If it should be overheated, an expansion will take place.

This steel will hold its size best when heated slowly and uniformly in a slightly oxidizing atmosphere to 1400 to 1440°F., soaked until uniformly heated throughout, and then quenched in oil. On small sizes the temperature can be as low as 1410°F., while on larger sizes it can be as high as 1440°F.; but, to get accuracy, the temperatures must be held within this range.

Decarburization.—This effect, the burning out of the carbon from the surface of the piece while in the hardening furnace, is of utmost importance in oil-hardening tool steel, because the tool is frequently not ground after hardening. The detrimental effect of a soft skin on a tool where accuracy is required is rather obvious.

Decarburization is caused by three things: (1) the analysis of the steel, (2) the temperature of the furnace, and (3) the furnace atmosphere. In regard to analysis, there are certain alloys, such as tungsten, that can be added to oil-hardening steel which make it more susceptible to decarburization. Since these alloys also raise the hardening temperature of the steel, this high heat will add to the effect. In this oil-hardening steel there are no such elements present, so that it is no more subject to decarburization than is a water-hardening steel.

Drawing.—The average Rockwell hardness values that may be expected at different drawing temperatures following an oil quench from 1425°F., are found in Table 11-XXI.

TABLE 11-XXI

Drawing Temperature 1 Hr. at Heat, °F.	Rockwell Hardness
Not drawn	C 64–65
200	64–65
300	63–64
350	62–63
375	61–62
400	60–61

The best combination of hardness and toughness in this type of steel is secured by drawing at 375 to 400°F. In larger sections, the draw should be longer than 1 hr. As a good rule, cut 25° off the drawing temperature every time you double the drawing time.

Uses.—When the requirements of the tool demand greater accuracy and greater safety in hardening, go west on the diagram and use this oil-hardening steel. It has found common application for broaches, blanking dies, cutter plates, gauges—both thread and ring—molding dies, thread-roller dies, master taps, spindles, trimming dies, can dies, etc.

Oil-hardening, Nonshrinkable Steel (Oil-tough).—Suppose that in the manufacture of a thread-roller die, the manganese oil-hardening steel just studied was used. This steel was applied, in this case, because it was necessary for the die to stay straight and flat during the hardening and for the pitch of the thread to remain accurate. However, in service it is found that the edges of the threads crumbled slightly and it is found that by drawing to about 600°F. this crumbling could be stopped. But the dies were then too soft and did not wear long enough. It is necessary, therefore, to change to a steel that is oil-hardening, that retains its nonshrinkable properties, and has increased toughness.

The steel to meet these requirements has the combined requirements of the steel placed at the west of our diagram and of the one placed at the south. Its position will be southwest and will have combined strength, toughness, and accuracy.

The typical analysis of this steel is as follows:

Elements	Per Cent
Carbon	0.75
Manganese	0.50
Silicon	0.25
Chromium	1.00
Nickel	1.75

The chromium and nickel in this case give the steel its toughness and wear resistance, and the carbon is sufficiently high to yield good tool hardness.

Heat-treatment. Annealing.—To anneal parts that have previously been hardened or annealed, it is recommended that they be packed in a suitable container with packing material, heated uniformly to 1400 to 1450°F., and cooled slowly in the furnace. For forgings, the treatment consists in an oil treatment from 1550°F., followed by an anneal from 1275 to 1300°F., to be followed by cooling slowly. To relieve machining strains for greater accuracy in hardening, the treatment consists in rough-machining, then annealing below the critical range, which is from 1275 to 1300°F., cooling slowly, and then finish-machining.

Hardening.—The hardening temperature range for this steel is 1475 to 1550°F. To secure the greatest accuracy in hardening it is recommended to harden from 1475 to 1500°F. If greater toughness and freedom from the hazard of cracking is desired, it is recommended to harden from 1525 to 1550°F.

Drawing.—The effect of draw temperatures on this steel are given in Table 12-XXI which shows the average Rockwell Hardness values at different drawing temperatures following an oil quench from 1500°F.

Table 12-XXI

Drawing Temperature 1 Hr. at Heat, °F.	Rockwell Hardness
As quenched	C 60–61
200	59–60
300	58–59
400	55–56

Experience has shown that the best drawing temperature to secure the maximum combination of hardness and toughness is 300°F. However, this steel has a useful drawing range at higher temperatures and, if a tool is sharply notched or of very irregular section, it may be drawn to 400 or even 500°F. The strength and hardness fall off rapidly at these higher draw temperatures, but there is an increase in ductility, which is sometimes necessary to meet very severe service conditions.

Uses.—This steel is used in three ways:

1. To replace the manganese oil-hard steel for those tools that require more toughness than the oil-hard steel possesses, since it is the tough member of the oil-hardening group. Thus, it finds its use in this connection for special taps, punches, blanking or forming dies, trimmer dies, thread-roller dies, forming rolls, embossing dies, etc.
2. Where hardening accuracy or cracking hazards make an oil-hardening steel necessary. Thus, it finds use in intricate shear blades, punches, stamps, swaging dies, slitting shears, etc.
3. For parts that are not tools but that in service require great strength and toughness combined with hardness and wear resistance. Such parts are clutch parts, heavy-duty lathe centers, dogs, stops, knuckle pins, spindles, clutch pins, etc.

High Chromium, High Carbon Steel (Oil-wear).—To complete the oil-hardening series of tool steels, we must, according to the diagram, use a steel that has the requisites of great wear resistance combined with safety and accuracy in hardening. Such a steel has the following typical analysis:

Elements	Per Cent
Carbon	2.10
Manganese	2.5
Silicon	0.25
Chromium	12.50
Nickel	0.50

The high carbon and chromium combine to form very hard iron-chromium carbides, which are closely distributed throughout the matrix. These hard carbides act as the wearing points and so give this steel its great resistance to abrasion. It has further qualities, in that it has the ability to withstand heavy pressures rather than shocks, hardens with very little change in size or shape, hardens to great depth, and, owing to the high percentage of chromium present, has mild corrosion-resistant properties.

Heat-treatment. Annealing.—This steel is annealed by packing in a suitable container with packing material, heating uniformly to 1500 to 1530°F., and then cooling very slowly in the furnace. The resultant Brinell hardness following this treatment ranges from 217 to 241. In this condition the steel may be machined without difficulty, although the tools cutting this grade should be quite sharp, owing to its work-hardening properties.

Hardening.—Table 13-XXI shows the effect of different hardening temperatures on this steel.

TABLE 13-XXI

Temperature, Oil-quenched from, °F.	Rockwell Hardness
1650	C 61–63
1700	C 64–65
1750	C 65–66
1800	C 65–66
1850	C 61–63
1950	C 54–56

From this table it can be seen that the maximum hardness is obtained only between the temperatures of 1750 and 1800°F. It is also found that the steel will hold its size best when hardened in this range. If it is hardened at temperatures above 1825°F., it gets softer, becomes somewhat nonmagnetic, and shrinks.

To harden properly it is recommended that the tools be packed in a suitable container with clean cast-iron borings (or charcoal mixed with 5 per cent NA_2CO_3), heated uniformly to 1750 to

1800°F., soaked at the heat at least 30 min., removed from the pack, and quenched in oil. The tool should be held in the bath until the temperature is below 200°F. Brittleness may result if it is tempered before reaching that temperature.

Drawing.—The effect of various drawing temperatures on this steel following an oil quench from 1775°F. is shown in Table 14-XXI.

TABLE 14-XXI

Drawing Temperature 1 Hr. at Heat, °F.	Rockwell Hardness
Not drawn	C 65–66
200	64–66
300	63–64
400	62–63
600	59–60
800	58–59
1000	51–53

The best drawing temperature to secure maximum hardness and wear resistance, with fair toughness, is found to be at 400°F. To obtain greater toughness at some sacrifice in hardness, a draw temperature of 800°F. should be used. The intermediate draw temperature of from 425 to 775°F. produces little gain in toughness.

Uses.—This steel is used for tools that demand a combination of hardening accuracy and safety with maximum wear resistance and greatest possible production. It is used to replace the oil-hard steel for tools that require exceptionally long life or that are designed for working abrasive materials. Such applications are broaches; slitting cutters; special taps; master tools; beading, forming and edging tools; subpress, forming, fender, thread-rolling, and lamination dies.

It is used to replace the water-wear steels for tools that require accuracy of size and shape, or in cases where the tool shape makes water quenching dangerous. Such applications are thread gauges, spinning tools, long mandrels, drawing dies, extrusion dies, reamers, sizing plugs, intricate punches, blanking dies, trimming dies, and special stamps.

High-speed Steel (Red-wear).—In the tool steels thus far studied we have been able to place steels in our diagram that would meet the demands of wear resistance, toughness, accuracy, and safety in hardening. There is one other property desirable in a tool steel and that is red-hardness. If a tool made from plain high carbon tool steel is heated in service above 300°F., it becomes softer and will fail to deliver the production that would be expected from a cold tool. Thus, we need a steel that will stay hard even though the source of heat (which may be either from friction or from machining hot steel) be sufficient to raise the temperature of the tool above the safe limit. Such a condition would, in the case of the carbon tool steel, draw the temper from the tool and leave it quite soft at the service heat.

Steels containing large amounts of tungsten and chromium do not lose their temper easily and retain much of their hardness even while they are hot. The following is a typical analysis of a so-called high-speed 18-4-1 type of tool steel.

Elements	Per Cent
Carbon	0.70
Manganese	0.25
Silicon	0.20
Chromium	3.75
Tungsten	18.50
Vanadium	1.10

Several variations from this analysis are the cobalt high-speed and the molybdenum high-speed steels.

The cobalt high-speed steels have a typical analysis as follows: carbon 0.77 per cent, tungsten 18.50 per cent, chromium 4.50 per cent, vanadium 1.75 per cent, molybdenum 1.00 per cent, and cobalt 9.00 per cent. The function of the cobalt is to improve the red-hardness quality, so that a cutting edge can be maintained under conditions of speed, feed, and depth of cut that would be impossible with the ordinary 18-4-1 high-speed steel. In opposition to these good qualities, cobalt has the disadvantage of reducing the toughness and forgeability of the steel, requires a higher hardening temperature than 18-4-1, and is somewhat more susceptible to decarburization.

The molybdenum high-speed steels have a typical analysis as follows: carbon 0.74 per cent, manganese 0.25 per cent, silicon 0.40 per cent, chromium 3.80 per cent, tungsten 1.50 per cent, molybdenum 8.50 per cent, and vanadium 1.10 per cent. This steel is a comparatively recent development and the belief is that it will find its field in high-speed cutting operations because of its machinability, toughness, and cutting qualities.

Solidification and Working.—It is apparent, now, that the essential alloying elements of high-speed steels are all alike (except cobalt) in that when they are alloyed with iron they tend to narrow the temperature range throughout which the gamma phase is stable, and, if present in sufficient amounts, will suppress this phase completely. In other words, all these elements when added to iron form a closed gamma loop. There is further similarity in that they are all essentially carbide-forming elements and, when present in the usual percentages, cannot be wholly dissolved by iron in the presence of a substantial amount of carbon.

Without delving into the intricacies of the equilibrium system of these steels let us discuss, in general, the solidification in an ingot of high-speed steel within the usual carbon range. In doing this let us consider the photomicrographs of the interior and exterior of an 8-in. high-speed steel ingot, shown in Fig. 17-XXI-*A* and *B*.

The first crystals to separate from such a melt will be gamma iron, with the various elements in solution. During this period and before reaching the eutectic temperature (there being both a eutectic and a eutectoid point, as in the iron-carbon equilibrium diagram) the austenite that forms will be coarse. The center of the grain, or point of initial solidification, will contain less of the alloying elements and carbon than will the last portions to solidify. This effect is shown by the differential attack of the etching reagent. At a temperature of 2380°F. (1300°C.) the last liquid melt to solidify will be the eutectic, which lies in the interstices between the crystals of coarse austenite. This massive structure is very similar to ledeburite of the iron-carbon alloys and is considered as composed of austenite and complex carbides. On further cooling, the proeutectoid carbide is rejected and is considered to be that portion in the photomicrograph that exists in thin plates or bands lying between the lightly etched martensitic zone and the portion of the austenite that transformed first. Its position is explained by a sluggishness in the alloy steels, which is not possessed by the carbon steels. That portion of the austenite that solidified last is rich in alloy content and lies adjacent to the eutectic areas. It rejects carbide, which finds a place in the zone next to the austenite, which is not expelling carbide because of its dilution. The austenite, which is quite rich, will

A *B*

FIG. 17-XXI.—Structure of the exterior *A* and interior *B* of an 8-in. high-speed steel ingot × 500. (*Courtesy of Vanadium-Alloy Steel Company.*)

transform to martensite upon reaching eutectoid temperature, this being comparable with the A_1 temperature of carbon steels. The eutectoid portion of the alloy, the center portion of the grain, is generally thoroughly spheroidized, as can be seen in Fig. 17-XXI-*A*. High-speed steels at room temperature will, therefore, contain three phases: ferrite, $(Fe, W, Cr, V)_6C$ and Fe_3C. The eutectic as shown in the photomicrographs has a structureless background, but with proper etching the background structure is made visible and is found to be decomposed austenite. The decompostition product will vary from spheroidized pearlite to fine lamellar pearlite, depending upon the rate at which the steel was cooled. The eutectoid area will also vary in structural characteristics with respect to the cooling rate.

The interior structure of the ingot differs somewhat from that of the exterior in the normal manner common to all cast sections. The grains in the interior are coarser and we find that the white eutectic regions are larger and more widely separated by the original austenite regions.

The elimination of the tungsten segregate is important to the production of the best high-speed steel structure. It is found that carbide out of solution in any form other than small spheroids is invariably detrimental. In ordinary carbon steels, the excess carbide is easily taken into solution by heating above the A_{cm} critical, but it is found that the usual anneal or hardening treatment will not break up this structure. The only method for breaking up this constituent is by the application of mechanical hot work. Hammering or so-called *cogging* will tear the segregate apart to a greater or less degree, while rolling simply elongates the segregate without materially breaking it up. Consequently cogging is a very important operation in the manufacture of high-speed steel. This working should be carried out at such temperatures and with such percentage reduction as will cause the massive tungsten segregate to be broken down and uniformly distributed throughout the matrix. The effect of such work on the distribution of the segregate is shown in Fig. 18-XXI-*A*, *B*, and *C*. The sections are all longitudinal, since a cross section would reveal only the ends of the tungsten segregate stringers and would not give a true picture of the distribution. Rolling in modern practice is mainly to obtain shape rather than effective breaking up of the segregate. Following proper working, the

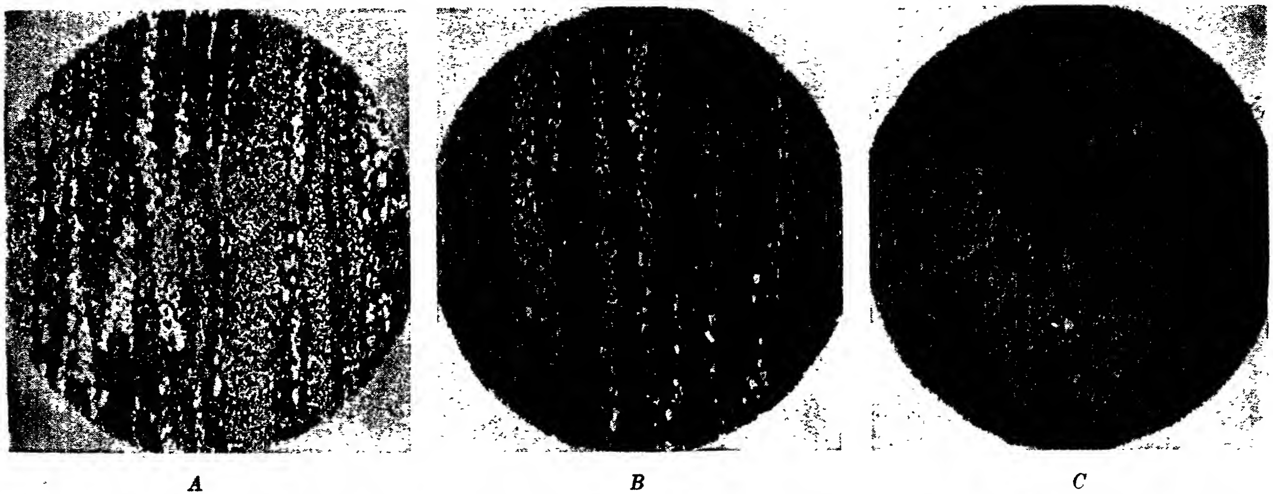

Fig. 18-XXI.—*A*, longitudinal section of a 3½ in. round from 8-in. ingot. Taken at × 500. *B*, longitudinal section of a 2-in. round from 8-in. ingot. Taken at × 500. *C*, longitudinal section of a ¾ in. round from 8-in. ingot. Taken at × 500. (*Courtesy of Vanadium-Alloy Steel Company.*)

small spheroidized particles precipitated from the solution will form a system of carbide particles distinct from the larger ones that resulted from the breaking up of the eutectic and proeutectoid carbide structures.

Annealing.—After being rolled or forged to billet form, the steel is annealed. This operation consists in heating to a temperature of around 1550 to 1650°F., holding for a sufficient length of time to effect partial solution of the alloys, and then cooling very slowly to below the critical range. A properly annealed bar will have a Brinell hardness of from 217 to 241. Figure 19-XXI shows a typical example of an annealed structure, indicating a matrix with more or less even-sized carbide particles.

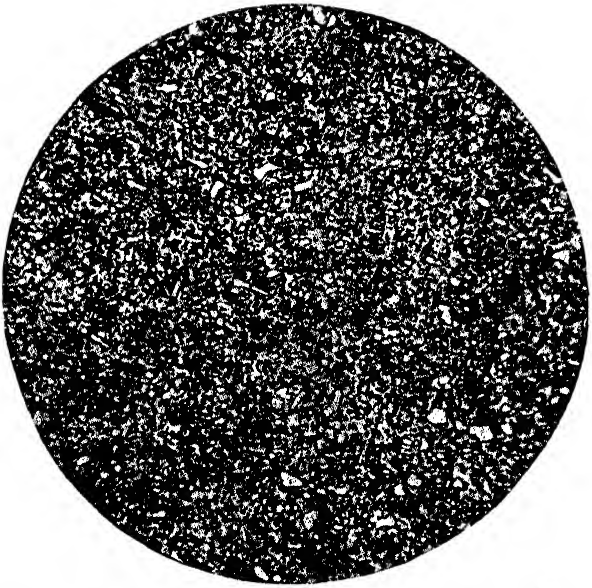

FIG. 19-XXI.—Structure of an annealed sample heated to 900°C. (1652°F.) cooled 7°C. per hr. to 650°C. (1202°F.). × 500. (*Courtesy of Carpenter Steel Company.*)

Hardening.—High-speed steel, as the name signifies, is essentially a cutting steel and can be used at high speeds without substantial loss of hardness. To produce the greatest cutting efficiency possible, it becomes necessary to harden the steel. The hardening of this steel is carried out by the same general processes as are used for ordinary carbon steels, with the formation of the same essential analogous constituents. The temperatures used are somewhat different, as has been observed in the study of the critical curves; and the composition is different, but this does not change the similarity of high-speed steel to hypereutectoid carbon steels.

The following practice assumes that the tools have been annealed after forging or if they are to be formed by machining, the bars as received are in the annealed condition. Final machining or preparation of the tool is done after hardening.

Heating for Hardening.—Heating for hardening has for its objective the solution of the tungsten-rich carbide particles to the greatest extent possible. The smaller particles, it will be found, will dissolve, but the larger particles will remain undissolved, regardless of the time and temperature involved. The appearance of the residual particles following heating and quenching is often a good index of the effectiveness of the heating for hardening. As could be assumed, the more carbides dissolved, the more the metal will be improved.

The time and temperature used for a given shape are best determined by experiment, but with the above conditions in mind an approximate cycle may be evolved. Obviously, where acute corners are present, we must heat very slowly and uniformly, for too steep a thermal gradient from outside to inside would result in overheating the corners and causing incipient fusion, excessive grain growth, and excessive oxidation. The recommended practice is to heat thoroughly, slowly, and uniformly to 1500 to 1550°F. (815 to 840°C.), and then transfer to a high-heat furnace held at a temperature of from 2250 to 2350°F., depending upon the distribution of the carbides and the size and shape of the tool; the character of the tools will determine the temperature used. The length of time in the high heat is important. It is usually the shortest possible period that is consistent with proper saturation, since holding at the high hardening temperatures would cause excessive grain growth and impart brittleness to the hardened tool.

Quenching.—After the tool has been heated to the proper temperature, it is usually quenched in oil or air. The difference in hardness obtained by using either quench is very small, usually not over two points Rockwell C hardness on large tools and less on small tools. Oil quenching is probably the more common practice of the two. In some cases of large intricate shapes where there is danger of cracking, air cooling is used, but here an added difficulty is encountered in the heavy scale that is formed and that must be removed. The steel should be cooled to a temperature definitely under the lower critical, which is approximately 750°F. If the steel has not been so cooled, the structure will not be changed by subsequent reheating or drawing.

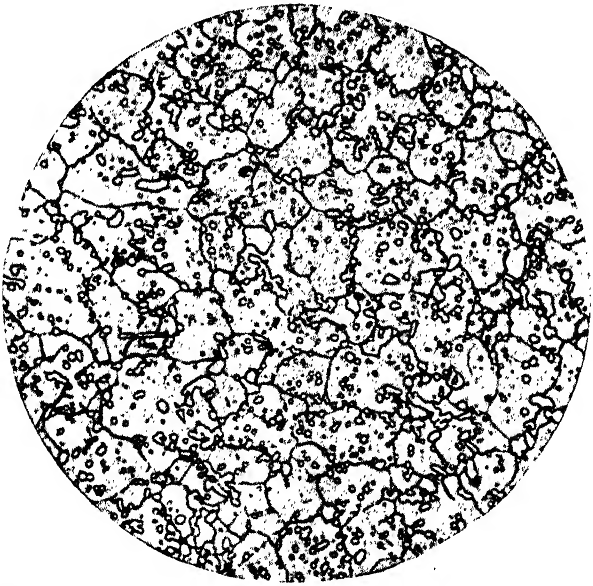

Fig. 20-XXI.—Structure of a sample oil treated from 2350°F. after holding 2 min. at heat. × 500. (*Courtesy of Carpenter Steel Company.*)

Drawing.—The steel should be reheated slowly and uniformly for tempering in either an open furnace, lead bath, or some suitable salt bath to a temperature of 1050 to 1150°F. The steel should be held at least 1 hr. at the temperature and then cooled in air.

The effect of hardening high-speed steel from a temperature of 2350°F. (1285°C.) not drawn, is shown in Fig. 20-XXI. The matrix in this specimen is quite resistant to attack by the etching solution and shows only traces of structure within the austenite grains. The structure is considered as composed of austenite, martensite, and undissolved complex carbides. Figure 21-XXI shows the same steel hardened at 2350°F. and drawn at 1050°F.

(565°C.). The matrix of this structure has undergone a marked change; the original austenite grains have become indistinct. This change is believed due to the breaking down of the austenite to martensite, and to the conversion of the martensite, originally formed, to troostite. The properties of this steel,[1] as quenched and drawn, will be described.

The curves indicate the following property trends: hardness, high, ranging from Rockwell C 62 to C 66; strength, very high; plasticity, high at the lower draws of the range but falling off sharply at the higher draws; toughness, very high in the vicinity of draw 900°F. (482°C.), then falling off to moderate values with higher draws.

Fig. 21-XXI.—Structure of a sample, oil treated from 2350°F. after holding 2 min. at heat and drawn at 1050°F. for 1½ hr. × 500. (*Courtesy of Carpenter Steel Company.*)

The effect of drawing temperatures on the Rockwell hardness number and coefficient of toughness (arbitrary combination of strength and plasticity) are shown in Fig. 22-XXI-*A* and *B*. The toughness coefficient indicates in a general way the ability

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[1] Emmons, J. V., Some Physical Properties of High Speed Steel, *Trans. A.S.S.T.*, **19**, 289–332 (1932).

of the steel to withstand rough treatment and service duty without breaking. Note should be made of the prominent peak in the vicinity of 900°F. (480°C.) and the distinct drop at higher draw temperatures. The peak indicates a high order of toughness at that particular draw temperature following a quench from 2350°F.

It is easily conceived that with different drawing temperatures we may be able to obtain structures varying from austenite—martensite, troostite, sorbite. These structures are all familiar, but the martensite thus formed in these steels exhibits several variations. In the as-quenched condition, the structure is believed to be composed largely of austenite with considerable martensite. The martensite formed on quenching may be called easily decomposable martensite. In quenched high-speed steel it is believed that we are dealing with martensite, which is easily converted to troostite at low-temperature draws, and with austenite, which is more stable and will become unstable only by drawing at high temperatures for a sufficient length of time. The result of the decomposition of the austenite results in the formation of a new "alpha" phase (alpha iron-carbide complex), which if it is compared with quenched and tempered carbon steel may be conceived as having little solubility for carbon; hence, carbides will be precipitated in a fine state of division and will, thus, add to the cutting properties of the steel. This phenome-

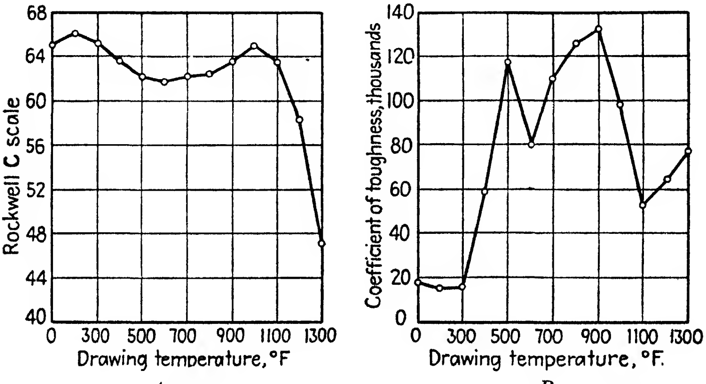

FIG. 22-XXI.—*A*, Rockwell hardness versus drawing temperatures hardened at 2350°F. (1285°C.). *B*, coefficient of toughness versus drawing temperatures hardened at 2350°F. (1285°C.) (*After J. V. Emmons.*)

non is described as secondary hardness. The effect of the decomposition on the hardness value is shown in Fig. 22-XXI-*A*. The peak of secondary hardness can be seen at about 1000°F.

The secondary-hardness phenomenon observed in high-speed steel does not differ in essence from a similar effect observed in the tempering of hardened carbon steels. The only difference is in the temperature of occurrence. The austenite of the high-speed steel is relatively stable and decomposes only at high temperatures, while the retained austenite of the carbon steels is more unstable and decomposes at much lower temperatures. In the hardened steel we also have martensite, which is unstable and undergoes decomposition. The reason for the decrease in toughness on drawing at temperatures exceeding 900°F., as is indicated by the curve, should now be evident. It should also be evident that martensite is the strong, as well as the hard, constituent of high-speed steel, troostite being the constituent that contributes the essential toughness, as martensite in itself has a low order of plasticity and toughness.

Red-hardness.—High-speed steel, as its name indicates, is essentially a cutting steel and can be used at high speeds without substantial loss in hardness. It will continue to cut readily when the cutting edge of the tool has become heated to a visible red heat. This property is known as the *red-hardness* of the steel. The cause of the red-hardness of high-speed steels is briefly as follows:[1] In hardened carbon tool steels the changes that cause the martensite to decompose or soften are due to the grain growth of the ferrite and the growth of the carbide particles above critical size. Similar changes take place in high-speed steels but only at a red heat. The reasons for the retention of fine grains in the ferrite of these steels are the increased resistance to growth due to the elements in atomic dispersion in the ferrite, and the abundance of obstructing carbide particles. The retention of critical-sized carbide particles is due to the great stability of the iron-tungsten carbide and the large size of the tungsten atom, which, owing to its size, will not diffuse into the ferrite space-lattice until a temperature corresponding to a red heat is reached.

Uses.—This steel has as its requirements maximum hardness and wear resistance for tools that get hotter than about 300°F. in

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[1] GROSSMAN and BAIN, "High Speed Steel," John Wiley & Sons, Inc., New York, 1931.

service. It, therefore, is the red-hard running mate of the water-wear steel previously discussed, and will have its position in the northeast part of the diagram. It is the only member of the red-hard group of steels suitable for making metal-cutting tools that become hot in service. Since everyone uses high-speed steel, it is the most important member of the group. One important·factor to be noted in using this steel is that at room temperatures the high-speed steel is no more wear resisting than is the water-wear steel.

This steel finds application for high-speed cutting tools, lathe tools, planer tools, drills, taps, reamers, broaches, milling cutters, form cutters, thread chasers, end mills, gear cutters, wood cutters, hot shear blades, hot punches, etc.

Low Carbon High-speed Steel (Red-hard).—In the forge shop a different set of conditions make their appearance. In this case we are dealing with a hot-working job, which demands that the steels maintain their hardness at elevated temperatures and that they resist abrasion, heat checking, shock or impact, and sudden temperature changes such as are brought about by water cooling in service. No single analysis will develop these characteristics to a degree that will make it universally suitable for all hot-work jobs. We know that the high carbon tool steel will not fit the job because such a tool becomes too hot in service. It, therefore, requires a red-hard tool, which can be secured by traveling east on the diagram.

The typical analysis for an all-round hot-working die steel for use in the forge shop is as follows:

Elements	Per Cent
Carbon	0.30
Manganese	0.30
Silicon	0.25
Chromium	4.00
Tungsten	14.50
Vanadium	0.50

The analysis shows this steel to be a low carbon high-speed steel in which the alloy content gives it red-hardness and wear resistance at elevated temperatures. The low carbon serves to balance this with sufficient toughness to make a good general-purpose hot-working die steel.

Heat-treatment. Annealing.—This steel is preferably annealed by packing in a suitable container with packing material, heating y and uniformly to 1550°F., holding at that temperature until the tool is thoroughly heated, and then slowly cooling in the furnace. Following this treatment, the steel will show a Brinell hardness of 197 to 223.

Hardening.—The hardening of this steel is brought about by following the same practice as that used for the red-wear steel. The tools should be preheated slowly and uniformly to 1500 to 1600°F., and thoroughly soaked at this temperature. Following the preheat, the tools are transferred to a superheating furnace at 2250 to 2300°F. When uniformly heated at this temperature, they are quenched in oil or in air when the resultant scale is not too objectionable.

The peak denoting secondary hardness can be observed as occurring at a draw temperature of 1100°F. At this peak the steel is just as hard as it was when quenched. Therefore, this steel should not be drawn lower than 1100°F. When greater toughness is desired, higher drawing temperatures than this should be used.

Drawing.—The effect of drawing temperatures on the steel following an oil quench from 2250°F. is shown in Table 15-XXI.

TABLE 15-XXI

Drawing temperature 1 hr. at heat, °F.	Rockwell hardness	Brinell hardness
As hardened	C—45	415
200	45	415
300	44	401
400	44	401
500	44	388
600	43	388
700	43	388
800	44	388
900	45	401
1000	45	401
1100	46	415
1200	45	401
1300	43	388

Uses.—The principal field for this steel is in hot-compression tools. It does not become hard enough to be used for metal-cutting tools and it lacks the extreme toughness needed in tools subjected to sharp hammer blows. This steel finds application for bolt and rivet dies, hot-drawing dies, hot-forming dies with shallow impressions, hot-piercing mandrels, hot-extrusion dies, forging mandrels, semihot-work dies, etc.

Low Chromium, Low Tungsten Hot-die Steel (Red-tough).—Now let us consider a job in which the hot-work die is used with a sharp hammer blow or contains deep recesses or sharp corners. The red-hard steel, previously discussed, cannot be made tough enough to meet these requirements even with a high draw. According to the diagram, greater toughness is obtained by traveling south and as the temperature encountered exceeds 300°F., the steel must also be red-hard. This, then, places the steel in the southeast part and completes the set of steels to be used at high temperatures.

The typical analysis of a steel that meets these requirements is as follows:

Elements	Per Cent
Carbon	0.55
Manganese	0.40
Silicon	0.30
Chromium	1.50
Tungsten	2.50
Vanadium	0.35

The analysis is so arranged that there is sufficient red-hard alloy to give the best red-hard properties without decreasing the great toughness that characterizes this steel.

This is a deep-hardening steel and it becomes much harder than the red-hard steel in the quenched condition but loses its hardness much more rapidly when used at elevated temperatures.

Heat-treatment. Annealing.—This steel is annealed by packing it in a suitable container with packing material, heating it uniformly to 1440 to 1475°F. and then cooling it slowly in the furnace. It will show a Brinell hardness, following this treatment, of 197 to 223.

Hardening.—Slow preheating is not so essential with this steel as it is with the higher tungsten tools. It is, however, good practice to preheat the tools on the forehearth of the furnace until they have acquired a dull-red color, then heat them to 1700°F., soak them for sufficient time, and quench them in oil.

Drawing.—Table 16-XXI shows the effect of drawing temperature on the steel after an oil quench from 1700°F.

TABLE 16-XXI

Drawing temperature 1 hr. at heat, °F.	Rockwell hardness	Brinell hardness
As hardened	C—61	653
200	60	627
300	59	627
400	57	601
500	55	578
600	54	555
700	52	534
800	51	514
900	48	477
1000	49	477
1100	48	461
1200	45	429
1300	35	321

It can be seen that there is slight evidence of secondary hardness at a drawing temperature of 1000°F. Experience has shown that the steel should seldom be drawn lower than this temperature. Drawing temperatures higher than 1000°F. can be used if experience indicates that this is necessary.

In making a comparison of hardness values of this steel with the red-hard steel, the high hardness value of this steel should not be misinterpreted. These high values are due to the fact that they are taken at room temperature, and that is not where the dies do their work. The red-hard steel is only slightly softened by high operating temperatures, while this steel is considerably softened. Thus, when the tools are operating at say 1000°F., the red-hard steel is harder than the red-tough steel. The wear resistance of the red-hard is also greater, owing to its higher alloy content.

Uses.—This steel finds use in hot-working tools which require the greatest possible toughness, particularly those subjected to heavy hammer blows or those containing deep recesses or sharp corners. For short jobs the steel may be used with economy for the tools for which the red-hard tool steel was recommended. This steel finds its application for heavy-duty compression tools, gripper or hot-forming dies with deep impressions, square bolt dies, bulldozer dies, hot header punches, hot shear blades having notched edges.

Suggested Questions for Study and Class Discussion

1. What special properties are obtained in steels containing over 25 per cent nickel?

2. How do chromium and nickel affect the corrosion resistance of a steel?

3. What type of steel would be applied in such high-temperature service as cracking still tubes or condenser tubes?

4. Describe the heat-treatment and the structure thus obtained for such applications. Why?

5. If these steels require resistance to oxidation at 1000°F., what alloy addition would you make to the regular analysis?

6. What is temper brittleness? What alloy would you add to the regular analysis effectively to eliminate this undesirable factor.

7. What element would you add to raise the creep strength? Why is creep important in such applications?

8. Why is it important to limit the chemical specifications so closely in a low carbon, 12 per cent chromium steel? Discuss this with respect to heat-treatment and structure produced.

9. Describe the manufacture of a carving knife from round stock; its processing, the heat-treatment.

10. Describe the structure of the round stock and that of the finished knife if the original cutlery-type steel were used. What difference would be noticed if the modified cutlery type were used?

11. What type of structure would you expect to observe in a 17 per cent chromium, 0.10 per cent carbon alloy? What difference would you expect in the structure of cast articles and finished strip?

12. What is "passivating"? In what way does it affect stainless steels?

13. What are some of the applications of 27 per cent chromium steels? What properties do they have that make them applicable in these services?

14. Why is 18-8 such an important alloy?

15. How is 18-8 annealed? In what way does this differ from our usual conception of annealing? Discuss the structural changes brought about by this treatment.

16. How is 18-8 hardened? What effect does such treatment have on the structure? Compare the effect of this treatment on the structure and properties with that on those of a carbon steel in the same carbon range.

17. What is the outstanding shortcoming of these higher chromium-nickel austenitic steels? What control means are used to eliminate this objectionable property? Discuss it from a structural standpoint.

18. What type of steel would apply in railroad crossings? In the races and balls of a large bearing? Will the same steel serve in both applications? Why?

19. What properties should a cutting tool possess?

20. List the typical types of tool steels and give the reasons why a carbon steel could not be applied in all cases.

21. What does the name "high-speed steel" signify?

22. Describe the heat-treatment of a high-speed steel tool and the structure produced after each treatment.

23. What is the phenomenon known as *secondary hardness*, as observed in the red-hard set of steels?

24. What is red-hardness?

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CHAPTER XXII

CAST IRONS AND THEIR TREATMENT

CLASSIFICATION

Cast irons may be generally classified in two ways. The first classification may be made according to the chemical composition. Cast irons are alloys of iron, carbon, and silicon containing from 1.7 to 4.5 per cent of carbon and 0.25 per cent or more of silicon, in combination with varying percentages of manganese, sulfur, and phosphorus, and sometimes one or more of the special alloying elements, such as nickel, chromium, molybdenum, vanadium, and titanium. Cast irons are sharply differentiated from steel in that they are not ordinarily malleable, as cast, at any temperature.

The second classification may be made according to the following conditions: (1) There are two outstanding factors that promote the formation of graphitic carbon, (*a*) a slow rate of cooling through and below the solidification range and (*b*) the presence of silicon; (2) there are two other outstanding factors or conditions that effectively promote the retention of carbon in its combined form, *i.e.*, (*a*) a fast rate of cooling through and below the solidification range and (*b*) the presence of much sulfur or manganese and little silicon or the addition of chromium.

With proper control of these factors, which essentially means a control of the structural condition of the carbon and the matrix, we may differentiate among three types of cast iron, according to the condition of the carbon present. This carbon may be (1) wholly in the graphitic condition, (2) wholly in the combined condition, and (3) partly graphitic and partly in the combined condition. Accordingly, we have the following broad commercial classification of irons: (1) gray iron, (2) white iron, and (3) chilled iron.

Gray Cast Irons.—These are irons having a composition in which a large portion of the carbon is distributed throughout the matrix as free or graphitic carbon in flake form. The name

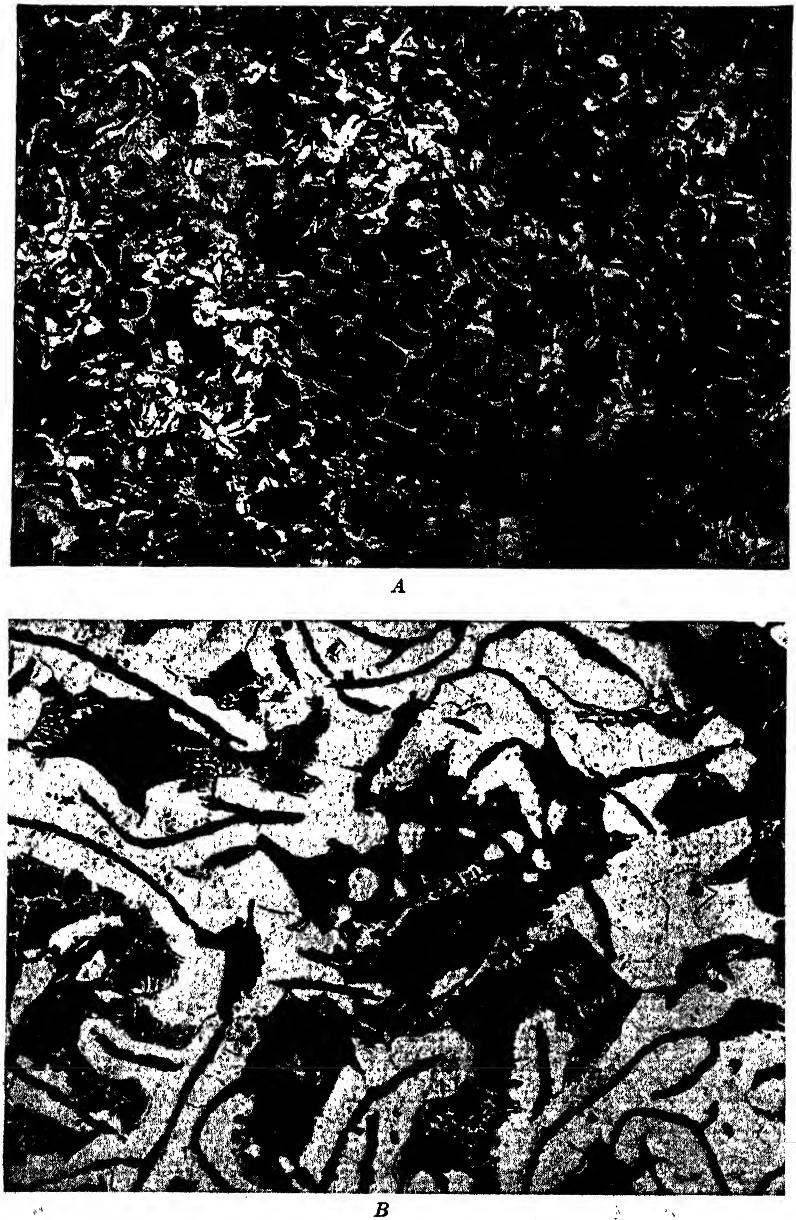

FIG. 1-XXII.—*A*, gray cast iron. Hypoeutectoid × 100 Nitric. *B*, gray cast iron. Hypoeutectoid × 500 Picric. (*Courtesy of National Malleable and Steel Casting Company.*)

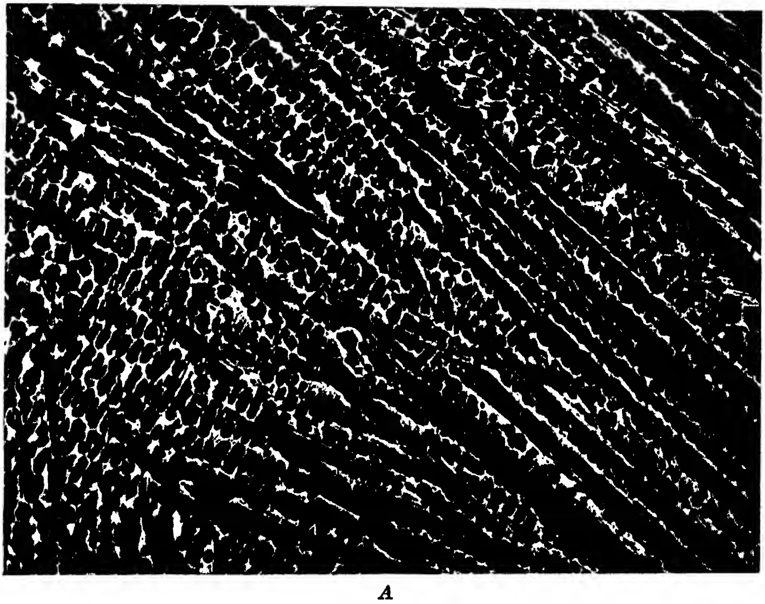

Fig. 2-XXII.—*A*, white cast iron. × 100 Nitric. *B*, white cast iron. × 500 Nitric. (*Courtesy of National Malleable and Steel Castings Company.*)

refers to the characteristic gray fracture of this iron (Fig. 1-XXII-*A* and *B*).

White Cast Irons.—The white irons are those that have all the carbon in the combined form as cementite or iron carbide (Fe_3C). The metal, when cast and broken, shows a fracture that is white in appearance; hence, the name. In this form the iron is extremely hard and brittle. (Fig. 2-XXII-*A* and *B*).

Chilled Cast Irons.—These are irons with some section purposely cooled so rapidly by means of chills that the carbon is retained in the combined form, and with the other section cooled normally and having the carbon in the form that is found in gray cast iron. Between these two sections lies a region of so-called *mottled* iron, in which some areas are white and others are gray.

INFLUENCE OF NORMAL ELEMENTS

Carbon.—Most commercial cast iron contains from 2.50 to 3.50 per cent of carbon as total carbon, which may exist either as graphite or free carbon, cementite or combined carbon, or the cementite lamellae of pearlite.

Graphite carbon in gray cast irons is present as thousands of curved, blackish-gray flakes, which are quite soft and weak and are directly responsible for the lack of ductility of cast irons, but which make it easily machinable. By interrupting the continuity of the structure they exert a very strong controlling effect upon the strength of the iron. Graphite flakes exert this effect according to their size, distribution, and amount. This is due to the fact that graphite, which has a low specific gravity, will occupy more volume than the matrix. Cast iron, for example, containing 3.00 per cent by weight of graphite, contains somewhere around 12 per cent by volume of that element.

The rate of cooling has a profound effect on both the shape and the size of the graphite particles, as well as upon the size of the grains of the matrix, and, subsequently, upon the physical properties. With castings of thin section, in which the cooling is rapid, the grains will be small; while with large castings, the rate of cooling will be relatively slow—hence, the grains will be larger, particularly toward the center. Large graphite flakes and a coarse grain structure are produced by a high silicon content, while dense, close-grained iron is produced by relatively low silicon content.

Silicon.—Silicon has a profound effect on cast irons, in that it tends to break down the combined carbon or cementite and is, therefore, the predominant element in determining the relative proportions of combined and graphitic carbon. All other factors being equal, the higher the silicon content the greater will be the relative amount of graphite and, consequently, the less combined carbon.

Thus, different types of cast irons, from the extremely hard white iron to the soft gray iron, can be obtained by controlling the silicon content. The amount of silicon used is dependent upon the physical requirements of each application and will range from 0.50 to 3.00 per cent.

The addition of alloying elements will modify the graphitizing effect of the silicon in the following manner: Elements that are carbide forming tend to reduce the graphitizing tendency, while those that graphitize will add to it.

Manganese.—The first action of manganese in cast irons is to combine with the major portion of the sulfur to form the compound manganese sulfide, which is relatively insoluble in molten and solid iron. This sulfide with other manganese compounds, such as manganese silicates, etc., appears in the structure as blue-gray or slate-colored inclusions.

The remaining manganese, *i.e.*, the manganese not present as inclusions, is considered as occurring as the carbide Mn_3C in irons containing combined carbon and is intimately mixed with the iron carbide Fe_3C. In those irons where the carbon is free or graphitic, the manganese occurs in solid solution in the matrix.

In ordinary commercial ranges the effect of manganese—which is to increase the stability of the carbide in irons—is not appreciable. When it is sufficient in amount to combine with the sulfur as MnS it promotes graphitization, for it tends to reduce the influence of the sulfur and allows the silicon full sway. If it is in excess of the amount needed to satisfy the sulfur, it retards graphitization because of the formation of Mn_3C or the double carbide $(FeMn)_3C$, which is more stable than Fe_3C alone.

In the ordinary commercial ranges (0.50 to 0.80 per cent), manganese has little effect on the physical properties of gray iron, this amount, in the main, inhibiting the effect of sulfur. When the additions exceed about 1.25 per cent, the strength, hardness, chill, and resistance to wear are increased and the machinability is decreased.

Sulfur.—Sulfur is present in steel largely as manganese sulfide and rarely as iron sulfide. In the absence of sufficient manganese, sulfur does increase the stability of the carbide and thus inhibits graphitization. This action is gradually decreased with an increase in silicon content. With sufficient manganese present (up to about 0.15 per cent), and under proper operating conditions, sulfur does not seem to exert a harmful effect on gray irons but it must be present in sufficient amount to prevent chill. In the absence of sufficient manganese, sulfur has a marked stabilizing action on the cementite and promotes chill.

Phosphorus.—Phosphorus is present in all cast irons, as an "impurity," in the structural component termed *Steadite*, which consists of iron phosphide (Fe_3P) dissolved in iron, and of iron phosphide. With normal percentages and under graphitizing conditions, phosphorus tends to promote graphitization, owing to the fact that it adds materially to the fluidity of the iron. This lengthens the time of solidification and, thus, favors cementite decomposition. Under conditions unfavorable to graphitization, however, a high phosphorus content will tend to retard cementite decomposition.

In the commercial cast irons, phosphorus is present in amounts ranging from about 0.15 to 0.90 per cent. Within this range there is no material effect on the tensile properties of the irons, but with amounts beyond this range, the Brinell hardness and stiffness are increased. Phosphorus, in some cases, acting through the "Steadite" component, improves wear resistance.

INFLUENCE OF ALLOYING ELEMENTS

Alloy cast irons are those irons that contain a specially added element or elements, as nickel, chromium, molybdenum, vanadium, copper, etc., in sufficient amounts to produce a measurable modification of the physical properties of the iron in the section under consideration. Irons to which special elements have been added should be used for special requirements and to meet certain physical specifications.

In general, a classification may be made of the alloying elements added to cast irons according to their effect on the micro-

TABLE 1-XXII.*—SUMMARY OF THE STRUCTURAL EFFECTS OF ALLOYS ON IRON

Elements	Per cent used in pearlitic irons	"Chill"	Effect on carbides at high temperature	Effect on graphite structure	Effect on combined carbon in pearlite	Effect on matrix
Chill-inducing						
Chromium	0.15–1.0	Increases[1]	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 and strengthens	Mildly increases	Refines pearlite and hardens
Mildly Chill-restraining						
Copper	0.50–2.0	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	0.10–3.0	Restrains[2]	Mildly decreases stability	Mildly refines	Mildly decreases and stabilizes at eutectoid	Refines pearlite and hardens
Titanium	0.15–0.10	Restrains	Decreases stability	Strongly refines[3]	Decreases	Produces ferrite and softens
Zirconium	0.10–0.30	Restrains		About neutral		Produces ferrite and softens

* From *Metal Progress*, **31**, 529 (1937).

[1] Chill-inducing effect about balances chill-restraining effect of 1½ parts of silicon or 2½ parts of nickel.

[2] Chill-restraining effect about ½ that of silicon.

[3] When added in small amounts and particularly when oxygen is also present.

structure. There are two principal effects: (1) on the amount, size, and shape of the graphite flakes; (2) on the amount of cementite, ferrite, and pearlite in the iron. Alloying elements also influence the fineness of the pearlitic structure. The total effect will, as a result, bring about considerable modification in the physical properties.

The alloying elements added serve to modify the matrix of the cast iron in essentially the same way as these same elements modify the structure of steel. Therefore, they will bring about changes in cast irons similar to those effected in steel.

A summary of the structural effects of alloys on cast iron is given in Table 1-XXII.

Nickel.—Nickel acts as a graphitizer in cast iron. In small amounts it will reduce the chill and, thereby, increase the machinability. The reduction of the chill tendencies is usually accomplished without a marked effect on the graphite structure or the grain of the gray iron. For general engineering applications, nickel is added to gray cast iron in amounts ranging from 0.25 to 2.0 per cent. It is also added in combination with other elements, such as chromium or molybdenum. When the nickel content exceeds 2.0 per cent, the irons become gradually harder so that within the range of 4.0 to 6.0 per cent this hardness property is utilized where the casting must withstand abrasive wear. When high nickel percentages are used, 12 per cent or more, in combination with other alloying elements, the irons are austenitic and have greater resistance to heat and corrosion than do ordinary irons.

The value of nickel in gray iron lies primarily in its effect on machinability. Nickel in gray iron produces castings that may be either harder or stronger and of better structure, with equal machinability in comparison with plain iron, or more readily machinable for the same hardness, strength, and structure.

Nickel, like other alloying elements, must be added in amounts that will bring about the most benefit after a study of the specifications of each individual case.

Chromium.—In contrast to nickel and silicon, chromium is a carbide-forming element. When added to cast iron, it forms a complex iron-chromium carbide, which is quite stable and will graphitize less readily than the straight iron carbides. When it is added in amounts exceeding about 0.80 per cent, it increases the combined carbon and intensifies the chill inclination of gray cast iron.

One of the most important uses of chromium additions to irons is where additional hardness or wear resistance is required over that of a given base iron. The chromium additions reduce the graphite content, decrease the flake size, and favor the formation of finer and harder pearlite. Added to a suitable base iron, chromium will add to the tensile strength, promote finer grain size, and increase the hardness. For general improvement of properties, chromium is used from 0.15 to 0.90 per cent with or without other alloying elements. Above about 1.00 per cent, free carbides appear and induce machining difficulties. At about 3.00 per cent chromium, the irons will show an entirely white fracture. Special-purpose white irons, containing 15 to 35 per cent chromium, have produced excellent results in improving heat and corrosion resistance.

Nickel and Chromium.—Nickel and chromium are used together in gray-iron castings for the purpose of refining the structure, increasing the hardness, obtaining greater wear resistance or strength, as required, without impairing the machinability.

There are a number of compositions of nickel-chromium irons produced for particular applications, but the range is generally not over 3.0 per cent nickel and 1.0 per cent chromium. The usual ratio of nickel to chromium for additions to gray iron is 2½ to 1, in which proportion they mutually neutralize each other with respect to chill. The alloy iron will then have the same chill tendencies as the plain iron to which it was added and the casting will be as machinable as the base iron. At the same time, the beneficial effects of both alloys will be retained and will be shown in the improvement in structure and properties of the casting.

Molybdenum.—Molybdenum additions to cast iron form both complex carbides and a solid solution with the ferrite. Its action as a carbide former is much less drastic than that of chromium. Molybdenum tends to prevent the agglomeration of free carbon and thereby causes the formation of finely divided graphite.

The molybdenum content usually ranges from 0.25 to 1.50 per cent. When it is used alone or in combination with other alloying elements in cast iron, it increases the tensile strength, hardness, and toughness. It is probably the most effective alloying element added to cast iron to increase the strength.

STRUCTURE OF CAST IRONS WITH RESPECT TO GRAPHITIZATION

Hypereutectic Alloys.—As such alloys start to solidify, the first constituent to separate is cementite, which continues to separate with decreasing temperature until the eutectic temperature is reached. At these high temperatures, cementite dissociates into graphite. This tendency is so great that even in the absence of silicon and with relatively rapid cooling it still takes place. At the eutectic temperature, eutectic cementite forms, which, with slow cooling and in the presence of the nuclei of the proeutectic graphite, easily decomposes to iron and graphite. In cooling from eutectic to eutectoid temperature, free or proeutectoid cementite is precipitated from the austenite. This cementite is less readily decomposed, even though assisted by the presence of the preexisting nuclei, because of the relatively low temperature now prevailing. Slow cooling and a large percentage of silicon would promote this decomposition. But if some of this proeutectoid cementite fails to be graphitized, we will have a hypereutectoid matrix and a probable so-called *mottle* structure (Fig. 3-XXII-*C*). If graphitization should be complete, a rare instance, the cast iron would then have a ferrite matrix.

Eutectic Alloys.—Upon reaching the solidification temperature, eutectic cementite is formed, which, if not inhibited by a fast rate of cooling, will decompose to austenite and graphite. On cooling from eutectic to eutectoid temperature, proeutectoid cementite is precipitated. The graphitization is promoted by the presence of the preexisting nuclei of eutectic graphite but is opposed by the low temperature. If this graphitization is incomplete, the matrix will be hypereutectoid (Fig. 3-XXII-*C*). Should the graphitization of the proeutectoid cementite be complete, we find that the eutectoid cementite formed at the eutectoid temperature is much less readily decomposed, owing to the low temperature. Its decomposition is generally incomplete and the resulting cast iron now has a hypoeutectoid matrix (Fig. 1-XXII-*B*).

Hypoeutectic Alloys.—In these alloys, austenite begins to separate at the point of solidification and, at the eutectic temperature, eutectic cementite is formed. This cementite decomposes

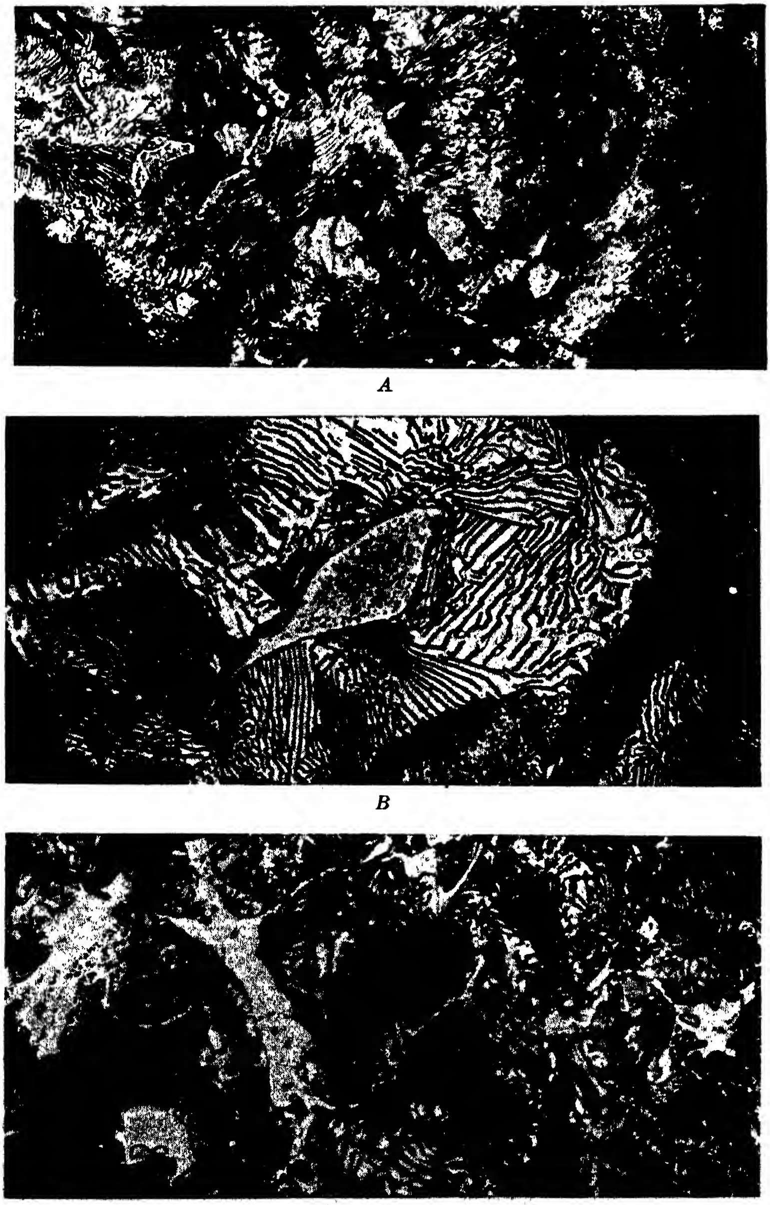

Fig. 3-XXII.—*A*, gray cast iron. Eutectoid × 500 Nitric. *B*, Steadite in eutectoid gray iron. × 500 Nitric. *C*, gray cast iron. Hypereutectoid × 500

quite readily, owing to the relatively high temperature. On cooling from the eutectic to eutectoid temperature, proeutectoid cementite is precipitated from the austenite and its graphitization proceeds the more slowly, as a result of the lower temperature of formation and the smaller assistance from the preexisting nuclei of graphite than in the previous two groups of alloys. The eutectoid cementite is graphitized only with great difficulty, owing to both the low temperature and the relatively small amount of preexisting graphite. If graphitization of the proeutectoid cementite is incomplete, the cast iron will have a hypereutectoid matrix (Fig. 3-XXII-*C*); while, with complete graphitization of the proeutectoid and incomplete graphitization of the eutectoid cementite, the matrix will be hypoeutectoid (Fig. 1-XXII-*B*). With complete graphitization of the proeutectoid cementite and no graphitization of the eutectoid cementite, the matrix will be eutectoid (Fig. 3-XXII-*A*).

Steadite.—This phosphorus-rich component in gray iron is largely composed of a binary cellular eutectic of iron and iron phosphide plus, in many cases, a structureless portion, which consists of solid solutions of iron and iron phosphide (Fig. 3-XXII-*B*). In the presence of combined carbon, *i.e.*, chilled and mottled irons, the phosphorus-rich component when rejected is a ternary eutectic of iron, iron phosphide, and cementite. During the graphitization, however, the cementite of the eutectic is rejected, leaving the binary eutectic. Where there is no combined carbon, phosphorus up to about 1.7 per cent is soluble in ferrite. Therefore, if we should heat a gray iron containing Steadite to 1800°F. and slowly cool—so-called *annealing*—the phosphorus would be found dissolved in the ferrite. In malleable irons (to be discussed later) Steadite is never found, for the above reason.

PHYSICAL PROPERTIES OF STRUCTURAL CONSTITUENTS

Since the engineering properties of a casting depend upon the structural make-up of the material and the degree of soundness of the material, a knowledge of structure is important. In Table 2-XXII the approximate structural compositions are shown for several typical irons. Types 1 to 5, inclusive, represent gray irons, while 6 is white iron and 7 is mottled iron.

An understanding and a correlation of these figures can be obtained by considering the properties of these compounds as they are shown in Table 3-XXII.

TABLE 2-XXII.—PROPORTIONS OF STRUCTURAL COMPONENTS IN SOME TYPICAL IRONS*

Type	Chemical composition, per cent						
	Graphite	Combined carbon	Silicon	Manganese	Phosphorus	Sulfur	Iron
1	2.10	0.80	1.10	0.75	0.20	0.07	94.98
2	2.50	0.70	1.80	0.80	0.32	0.10	93.78
3	2.65	0.65	1.25	1.00	0.10	0.12	94.23
4	2.50	0.72	1.50	0.90	0.30	0.11	93.97
5	3.09	0.40	2.40	0.55	1.05	0.10	92.41
6	....	3.30	0.60	0.52	0.50	0.15	94.93
7	1.50	1.80	0.92	0.36	0.22	0.13	95.07

Type	Structural composition, per cent by volume							
	Silicon ferrite	Pearlite	Manganese sulfide	Steadite	Graphite	Iron sulfide	Free cementite	Calculated specific gravity
1	6.36	84.59	0.35	2.00	6.70	....	.....	7.34
2	15.65	72.84	0.49	3.15	7.87	....	.....	7.23
3	21.98	67.89	0.60	0.98	8.35	....	.....	7.25
4	12.98	75.64	0.54	2.96	7.88	....	.....	7.25
5	39.14	40.75	0.48	10.12	9.51	....	.....	7.08
6	2.23	49.13	0.67	5.21	....	0.09	42.67	7.66
7	3.30	72.71	0.45	2.23	4.84	0.21	16.25	7.43

* AMERICAN FOUNDRYMEN'S ASSOCIATION, "Cast Metals Handbook," Chicago, 1940.

Mass and Cooling Rate.—When reference is made to the influence of mass or section size in cast iron, what is really meant is cooling rate. It is known that the more slowly a given gray-iron casting is cooled from the time of pouring, the lower will be its strength and hardness. Furthermore, the more slowly a given

TABLE 3-XXII.—PROPERTIES OF COMPOUNDS AND MICROCONSTITUENTS OF CAST IRON*

Constituents	Specific gravity	Tensile strength, p.s.i.	Elongation in 2 in., per cent	Brinell hardness	Remarks
Ferrite	7.87	40,000 ±	40	90	Iron
Silicoferrite	.....	45,150	50	88	Contains 0.82 per cent Si
Silicoferrite	.....	63,500	50	124	Contains 2.28 per cent Si
Silicoferrite	.....	77,400	21	150	Contains 3.40 per cent Si
Pearlite	7.846	120,000	15	240	Laminated structure containing 6½ ferrite to 1 cementite
Cementite	7.66	5,000	..	550	Compound of iron and carbon containing 6.67 per cent carbon
Graphite	2.55	....	..	....	
Ledeburite	.....	....	..	680–840	Massive cementite plus saturated austenite
Steadite	7.32	....	Brittle	....	Phosphide of iron (Fe_3P) and saturated solution of Fe_3P in iron. This substance occurs by volume 10 times that of phosphorus by weight
Manganese sulfide	4.00	....	Brittle	....	1.73 parts Mn to 1 part S
Iron sulfide	5.02	....	Brittle	....	1.75 parts Fe to 1 part S

* AMERICAN FOUNDRYMEN'S ASSOCIATION, "Cast Metals Handbook," p. 340, Chicago, 1940.

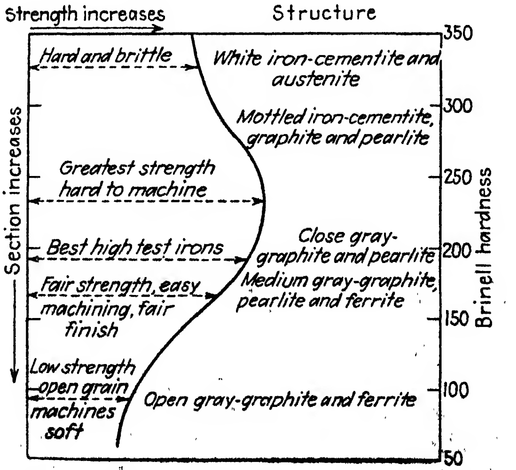

FIG. 4-XXII.—The general interrelationship between structure, cooling rate, and physical properties of cast iron. (*From the* 1940 *Cast Metals Handbook.*)

iron is cooled through the graphitization range, the more complete will be the graphitization. There will also be greater opportunity for grain growth, variation in the degree lamination of the pearlite, and increase in flake size of the graphite carbon.

The general relationship of structure, strength, hardness, and section size is shown in Fig. 4-XXII.

The relationship is quite distinct and as would be expected. In other words, the ferritic irons, having less of the strong pearlite constituent and more of the weakening graphite, are not so strong as irons that have a fully pearlitic matrix, provided that the total carbon and section size are the same. A pearlitic matrix is very desirable, and most high-strength irons have such a matrix. A pearlitic matrix, however, does not guarantee maximum physical properties, as the amount and condition of the graphite play predominant parts in controlling these conditions.

PROPERTIES OF CAST IRON

The following discussion will serve only as a summary of some of the properties of cast irons. For a more complete discussion, reference should be made to the bibliography at the end of this chapter, from which the following material is taken.

Tensile Strength.—For gray cast irons the tensile strength ranges from less than 20,000 p.s.i. for soft-weak irons to over 70,000 p.s.i. for some of the high-strength and heat-treated irons, especially if alloyed.

Compressive Strength.—This property, along with ease of machining and low cost of production, is the outstanding characteristic of cast iron. Because of this property, frames and bases of machine tools and equipment of many kinds are commonly made of cast iron. The compressive strength is about three to five times the tensile strength, the narrower ratio applying to the stronger irons and the wide ratio to the weaker irons. For gray irons the range would be from about 70,000 to 200,000 p.s.i.

Transverse Strength.—The transverse test is the one most commonly used for cast iron. Its principal use is as a foundry control test. In the test, the cast test bar is supported on two knife edges, whose distance apart is known as the *span*, and is broken by a load applied midway between supports. The resulting deflection is reported in inches. Since the load and the deflection depend upon the span and the size of the section, as well as upon the material, the size of the test bars must be specified. Three sizes of test bars are specified by A.S.T.M., to represent various thicknesses of castings and thereby assume various cooling rates. An agreement must be made in the specifications on the controlling thickness of the casting, so that the corresponding bar can be used.

Hardness.—The hardness range obtainable in cast iron varies from 130 to 210 Brinell in the as-cast commercial gray irons to approximately 450 for ordinary white irons. The alloy gray cast irons will run higher than the above figure and heat-treated irons may go above 600 Brinell. Alloy white irons may run above 600.

Machinability.—Machinability in cast irons[1] varies according to their microstructure in the following order:

1. Ferritic (free carbon and silicoferrite).
2. Pearlite–ferritic (free carbon, silicoferrite, and pearlite).
3. Pearlitic (free carbon and pearlite).
4. Mottled iron (pearlite, free carbon, and enough massive cementite to produce a mottled fracture).
5. White iron (pearlite and massive cementite).

The machinability, as can be seen, is dependent upon the arrangement of the hard and soft structural components. The ferritic irons are the easiest to machine because the carbon is all present as finely divided free graphite flakes. At the other extreme, the chilled or white iron is difficult to machine because the carbon is present both as massive free cementite and as the carbide in the pearlite with no graphite flakes.

Wear.—Wear resistance of gray irons at various temperatures is usually good, because they are exceptionally free from the tendency to gall or seize. This is due to the fact that the graphite flakes produce a self-lubricating action. Under conditions where resistance to abrasion is required and where there is a high unit-load application and relative motion between the moving members, white or chilled cast iron is superior, on account of its hardness, which is a necessary prerequisite to prevent indentation.

Laboratory tests on conditions of wear in cast irons have shown the following conclusions:[2]

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[1] AMERICAN FOUNDRYMEN'S ASSOCIATION, "Cast Metals Handbook," p. 382, Chicago, 1940.

[2] A.S.M. "Metals Handbook," p. 634, Cleveland, Ohio, 1939.

1. Wear resistance increases with matrix hardness.
2. With increasing combined carbon content, the resistance to wear increases rapidly up to 0.60 per cent combined carbon and then slightly up to 0.85 per cent.
3. Permanent-mold cast iron wears faster than sand-cast iron of the same combined carbon content.
4. Increasing silicon content decreases resistance to wear.

Growth.—The term "growth," as applied in cast irons, refers to the increase in volume that sometimes occurs when the metal is repeatedly heated and cooled, or when it is held at high temperature. This change in volume appears to be due principally to graphitization of the combined carbon and the penetration of oxidizing gases along particles of graphite, which causes oxidation of ferrite. These effects result in the formation of fine cracks and mechanical swelling and cause the castings to become weak, brittle, and misshapen. At temperatures below 800°F., in air, comparatively little trouble is encountered from growth. However, it may become a factor at lower temperatures than this if the iron is exposed to certain corrosive influences, chiefly oxidation. As the temperature is increased beyond 800°F., growth becomes greater and much more rapid.

The tendency of cast iron to grow is directly dependent upon the silicon content or content of other graphitizing elements. Irons most favorable to growth resistance are those having stable carbides and close, dense grain structure. Therefore, the elements that stabilize carbides and produce close grain structure retard growth. The irons found to be advantageous in resisting growth are low silicon irons; irons containing chromium (0.30 to 1.00 per cent); high silicon iron (6 to 8 per cent); and special irons, of which the composition containing 14 per cent or more nickel, about 5 per cent copper, and 1 to 4 per cent chromium is probably the most widely used. This type of special iron is austenitic.

HEAT-TREATMENT

Cast iron reacts to heat-treatment in a manner comparable to steel, since the only fundamental difference in the composition of the two materials is the presence of free graphite particles in gray cast iron. The heat-treatment of cast irons will, for our present purpose, be divided into two classifications: (1) annealing and (2) quenching and drawing.

Annealing.—The heat-treatment used for the relief of internal stresses is a low-temperature annealing treatment carried out at a temperature of between 850 and 950°F. This treatment, which is referred to as *mild annealing*, *normalizing*, and *aging*, is employed to prevent distortion after machining. In former years the same treatment was carried out by allowing the castings to "age" for many months at ordinary temperatures prior to machining, this aging process bringing about the relief of the internal stresses. The low-temperature anneal treatment now accomplishes the same purpose in a few hours.

The time of holding at the annealing temperature will depend upon the size of the section. This time will range from 30 min. to 5 hr., after which the casting is usually slowly cooled in the furnace. This treatment will result in very little decomposition of cementite and only slight change in the hardness or strength properties of the castings.

It is often necessary to soften castings that contain some free cementite, so that rapid machining may be facilitated. This treatment is generally carried out in the temperature range from 1400 to 1500°F., although it may be quite satisfactorily carried out at somewhat lower temperatures, with an increase in time. In all annealing treatments, extreme care should be taken to prevent oxidation of the castings.

In general, it has been found that annealing decreases the strength and amount of combined carbon and increases the graphite content. The length of time at the temperature and maximum temperature used are the determining factors affecting the changes for an iron of a given composition. Low silicon, low total carbon irons will resist the structural changes more than a high silicon, high total carbon iron. For the former a higher temperature and longer length of time at the temperature must be used to obtain the same degree of softening.

Fully annealed plain gray irons may have a Brinell hardness of from 135 to 155, while completely annealed alloy irons may have a Brinell hardness of from 140 to 180 or higher, depending upon the composition.

Quenching and Drawing.—The use of a quenching and drawing on gray cast iron is not altogether new, but only comparatively recently has it received commercial application. Castings of this type subjected to a quench and draw will have greater hardness and strength and will, as a result, have increased wear and abrasive resistance.

In general, the procedure followed and the precautions to be taken in quenching and drawing cast irons are the same as those used in the treatment of carbon and alloy steel sections. The castings should be heated quite slowly to the quenching temperature, which will range, for most of the irons used in production at the present time, from 1450 to 1550°F., after which they are held for sufficient length of time for complete heat penetration, and quenched, usually in oil, and tempered somewhere in the range of 450 to 1000°F., depending upon the toughness required.

As in the case of steel, the hardenability of a particular iron is dependent upon the amount of carbon and other elements in solution. In the plain carbon irons, carbon, which, as we know, is a very important factor in the hardening power, will generally be present in an amount ranging from 2.00 to 4.00 per cent. Most of this will be present in the form of inert graphite flakes. It is plainly seen that, if the carbon is present in large, interlocking graphite flakes with a ferritic matrix, it will not be materially improved by a hardening treatment. Therefore, the ungraphitized carbon and the size, shape, and distribution of the graphite particles are the important factors in the hardening of irons.

The most common method for increasing hardenability would be, then, to increase the amount of dissolved or combined carbon. A correctly balanced iron[1] should contain a combined carbon content of from 0.60 to 0.80 per cent, with none of the free carbon segregated into knots or clusters. A *correctly balanced* iron is, therefore, defined as containing just enough combined carbon to yield a pearlitic structure; an *overbalanced* iron is one with excessive graphitization; an *underbalanced* iron, one with inadequate graphitization. We shall assume that a balanced composition is being used, one in which the graphite flakes are small and well distributed and in which the body is fully pearlitic. This type of iron will be low in silicon, which will, as we know, reduce the tendency to graphitize; it will be initially finer grained, and will approach the desired hardness value of 170 to 190 Brinell.

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[1] VANICK, J. S., Hardening Cast Iron with Alloy Additions, *Metal Progress*, **28** (No. 6), 42–47, 80 (1935).

Owing to the high carbon content, the critical range will be low, which means that the quenching temperature range will be low. Where the alloying elements add up to less than 5 per cent, the correct quenching range is 1450 to 1550°F.

With so high a carbon content, it is quite apparent that the iron is strongly susceptible to cracking if very rapid rates of cooling are applied. Oil quenching is apparently very satisfactory under these conditions.

Following the quench, the castings are tempered. Tempering is, preferably, started while the castings are still warm and consists in reheating slowly to 400 to 1000°F., followed by slow cooling or with the furnace.

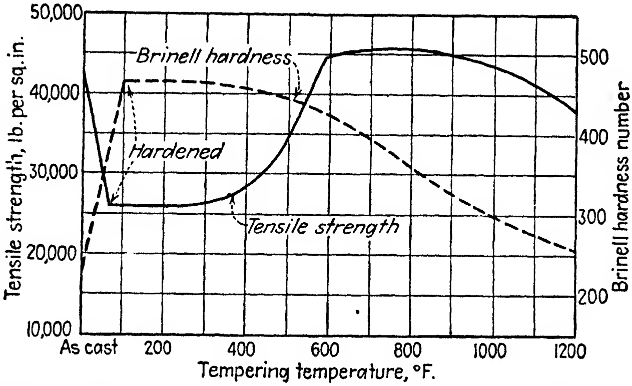

FIG. 5-XXII.—Effect of tempering temperatures on the properties of cast iron. (*From* 1940 *Cast Metals Handbook*.)

In the hardened, as-quenched condition, the strength figures are materially altered. The ultimate breaking strength being materially reduced. The stress-strain curves show a marked change in character, in that the characteristic curvature almost completely disappears, the curve becoming practically truly proportional right up to the breaking strength, with complete absence of any plasticity. The hardness accompanying this change increases to within the neighborhood of 450 to 500 Brinell.

Upon the tempering of this hardened iron at progressively increasing temperatures, a very interesting effect is noted. At temperatures as low as 100°C. (212°F.) there is a noticeable increase in the ultimate strength. With increasing temperature, the ultimate strength continues to increase and the appearance of the stress-strain curve reverts to the more normal type, showing distinct curvature in the higher stress ranges to the ultimate breaking strength. Figure 5-XXII shows the effect of quenching and drawing on the hardness and strength of a particular iron. It shows that tempering from 500 to 1000°F. brings about an increase in tensile and impact properties, which are superior to those of the material in the as-cast condition. Investigation has shown that, with higher tempering temperatures, this maximum value for mechanical properties will fall off and will approach the values of the original as-cast condition.

Machining operations on castings must be carried out prior to heat-treatment. In order that the machining operations may be speeded up, the castings are frequently given an annealing treatment and then hardened and tempered as desired. Examples have been shown where an iron as cast had a Brinell of 280, was annealed to 180, machined, then quenched to 500 and tempered. The proper cycle of treatment must, of course, be observed and attention must also be given to volume changes that take place. Both annealing, and quenching and drawing, will cause such changes, but these can be taken out by final grinding if close tolerances are required.

MALLEABLE CAST IRON

The production of malleable cast iron takes place in two steps. The first is a matter of producing a hard, brittle white-iron casting, which, as we know, is primarily an alloy of iron in which there is carbon, all in the combined form, silicon, as an iron silicide, and a small amount of other elements. Structurally it consists of massive cementite and pearlite. The second step consists in converting, through a proper heat-treating cycle, the hard and brittle castings into a product that is easily machinable and, although soft and ductile, quite tough and strong. This step, which may be termed an *annealing* process, breaks down the combined carbon and causes it to precipitate as free carbon or graphite in a matrix of ferrite. The graphite, in the process, has collected in nodules called *temper carbon*, or *secondary graphite*, in the matrix.

Annealing.—The annealing process used in the production of malleable iron consists of two stages. The first stage consists in heating to a temperature well above the critical range until the massive cementite has been destroyed. The commercial temperatures range from 1600 to 1700°F. (870 to 925°C.). The usual changes take place during this heating; the pearlite transforms to austenite; some of the free cementite goes into solution in the austenite, in an amount equivalent to that required to saturate the austenite at that temperature; and, the carbon or cementite above that required for saturation precipitates as free carbon or "temper carbon."

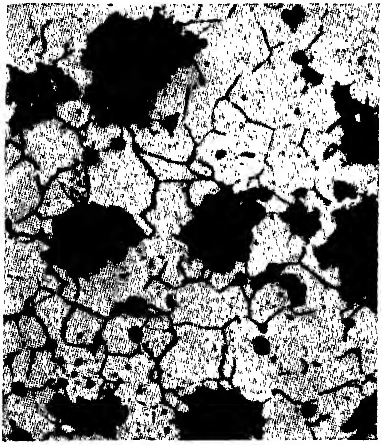

FIG. 6-XXII.—Air furnace malleable iron made by standard long cycle anneal (140 hr.). Note large graphite particles and equally large average grain size. Etched × 100.

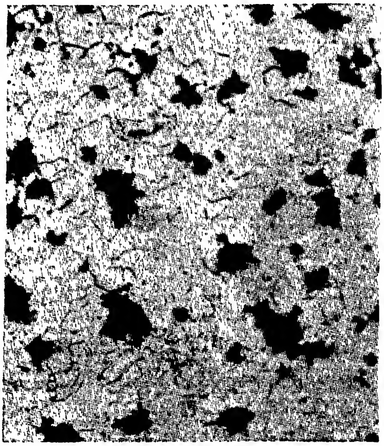

FIG. 7-XXII.—Malleable iron annealed with 20 hr. cycle, having fine graphite nodules, uniformly distributed and fine grain size. × 100. White iron was melted in an indirect arc rocking electric furnace.

(*Courtesy of American Society for Metals.*)

After equilibrium at the given temperature has been reached, the iron is slowly cooled. Since the solubility of carbon in austenite decreases as the temperature is lowered, carbon will be precipitated at nuclei that have formed at the soaking temperature. This nuclear formation is an inherent capacity of the iron which is set in motion by the heat energy supplied in annealing. When the temperature in cooling has approximately reached the lower critical, it will be composed of the carbon that has collected in nodule form in pearlite.

The second stage of the process consists of holding at a temperature just below the lower critical point (Ar_1) so that the cementite of the pearlite will be spheroidized and then be graphitized.

The temperatures used range from 1300 to 1350°F. (705 to 730°C.). When equilibrium is reached, the structure is composed of temper carbon in a matrix of ferrite (Fig. 6-XXII).

From the previous discussion it is apparent that malleable iron of acceptable quality should consist entirely of ferrite and temper carbon. To obtain this result, the requirements are that the composition of the melt shall be kept within certain limits, particularly with regard to carbon and silicon, and that the original casting shall be entirely free of flaky graphite. The cross section of the casting must also be considered, as it is an important factor in determining the rate of cooling in the molds, which directly affect the amount of carbon that will separate. In general, to secure the most desirable properties, the procedure is to keep the silicon low with high carbon iron, and high in low carbon irons. Silicon may be as much as 1.20 to 1.30 per cent in small, low carbon castings or as low as 0.50 to 0.60 per cent in large, high carbon castings.

A typical cycle[1] for producing malleable iron in batch-type annealers is as follows:

	Hours
1. Heat to 1650°F. (900°C.)	20
2. Hold at 1650°F. (900°C.)	50
3. Cool to 1250°F. (675°C.)	70
4. Cool to handling	5

The properties of the malleable iron are dependent to a large extent upon the composition of the white iron, but for a more or less standardized composition of white iron produced by air furnace melting practice, the analysis:

Elements	Per Cent
Carbon	2.40
Silicon	1.10
Phosphorus	0.15
Sulfur	0.05
Manganese	0.25

will produce, after annealing, the following physical properties:

Ultimate tensile strength	50,000–60,000 p.s.i.
Yield point	30,000–40,000 p.s.i.
Elongation in 2 in.	15–20 per cent
Impact strength, Charpy	7–10 ft.-lb.

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[1] Morken, C. H., The Heat Treatment of Cast Iron, *Trans. A.S.M.*, **22**, 227–248 (1934).

Short Cycle.—Owing to the length of the annealing cycle, considerable investigation has been carried out in an effort to shorten it. Particular attention has been given to the type of annealing furnace used, and it is found that by the use of continuous-type furnaces the annealing time can be very effectively reduced. For example, with an iron of around 2.75 per cent carbon and 1.15 per cent silicon, annealing in a tunnel-type furnace, the iron will require 50 hr.; while in a batch-type furnace it would require from 90 to 110 hr. to accomplish the same result.

According to studies by White and Schneidewind[1] the factors that tend to reduce the annealing time are brought about by altering the size, distribution, and number of particles of temper carbon. Graphitizing is obviously a diffusion process, the combined carbon diffusing to and depositing upon a graphite nucleus. If these graphite nuclei are close together, it means that the diffusion distance for any atom of carbon has been decreased and graphitization will be speeded up. Malleable iron produced in a short cycle should be one in which the graphite nuclei have rapidly formed in large numbers.

The two investigators just mentioned and others have studied the influence of superheating the molten white iron. It has been found that white iron heated to 3000 to 3200°F. will anneal in 40 per cent of the time required to anneal iron that has been superheated to 2800°F. There seem to be several factors that contribute toward this effect. Evidently, since these irons were made in electric furnaces, the combination of thorough mixing with superheating distributed the free cementite in such a way that several graphitization centers existed where only one would exist in white iron produced in the usual manner.

Another advantage of superheated white iron, when melted under reducing conditions in a rocking, indirect-arc furnace, is its ability to contain abnormally high quantities of silicon without the formation of primary graphite. The massive cementite is found to be uniformly distributed throughout the pearlitic matrix.

A typical white iron produced in this manner runs about 2.40 per cent carbon, 1.65 per cent silicon, low phosphorus, sulfur, and

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[1] WHITE and SCHNEIDEWIND, Malleable Cast Iron Annealed in Short Cycle, *Metal Progress*, **25** (No. 6), 25–29 (1934).

manganese. A typical cycle for lots of from 1500 lb. to 5 tons is as follows (Morken):

	Hours
1. Heat to 1750°F. (954°C.)	3
2. Hold at 1750°F. (954°C.)	4
3. Cool to 1250°F. (677°C.)	14

The following physical properties are attained.

Ultimate tensile strength	60,000 p.s.i.
Yield point	40,000 p.s.i.
Elongation in 2 in	15 per cent
Impact strength, Charpy	7 ft.-lb.
B.H.N	150

Figure 7-XXII is a typical structure of malleable iron made in the above manner. It is interesting to compare the difference in grain size, carbon distribution, and particle size with the structure obtained in the conventional process (Fig. 6-XXII).

Special Irons (Pearlitic Malleable Irons).—Recently, processes have been developed whereby white irons are given a wide variety of special heat-treatments. These treatments all start by holding the castings at high temperatures, ranging from 1700 to 1800°F. (925 to 980°C.), to decompose the free cementite. Following this, the heat-treatment may be varied in a number of ways, much the same as for steel. These products differ from steel only in that they contain temper carbon.

One effective treatment for sections rather simple in design is to use an oil quench following the high-temperature heat. This will produce a martensitic structure with high strength. An air quench, which is less severe, followed by a draw, is also quite effective.

One treatment is as follows: (Morken):

	Hours
1. Hold at 1750°F. (954°C.)	3
2. Air quench to 1400°F. (760°C.)	
3. Oil quench	
4. Draw at 1350°F. (732°C.)	1
5. Air quench for handling	

The physical properties of this type average as follows:

Ultimate tensile strength	100,000–125,000 p.s.i.
Yield point	85,000–100,000 p.s.i.
Elongation in 2 in	3–7 per cent
Endurance limit	40–50 per cent
Impact strength (Charpy)	5–7 ft.-lb.
B.H.N	225–275

Figure 8-XXII shows the structure obtained by this treatment. This structure probably is sorbitic.

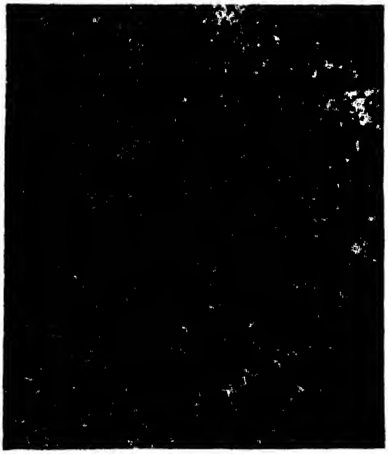

FIG. 8-XXII.—Iron annealed with 24 hr. cycle, but heat treated to produce high strength with a total cycle of 8 hr. Etched × 100.

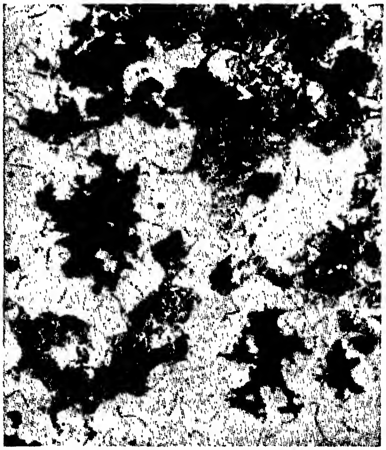

FIG. 9-XXII.—Showing retention of pearlite. Etched × 100.

(*Courtesy of American Society for Metals.*)

There are, of course, many other treatments intermediate between the complete anneal and the quench-draw treatment. These treatments produce, in some cases, a pearlitic or sorbitic matrix and, in other cases, a pearlitic and ferritic matrix. By proper treatment the relative proportions of these phases can be controlled to obtain a given set of properties.

A fully pearlitic matrix does not give properties as good as the above high-strength iron, so most producers arrest at the low-temperature critical long enough to produce some ferrite, to increase the ductility. Figure 9-XXII illustrates such an iron produced with the following cycle (Morken):

	Hours
1. Hold at 1750°F. (955°C.)	3
2. Cool in the furnace to 1350°F. (730°C.)	
3. Hold at 1350°F. (730°C.)	5
4. Cool in furnace at 1200°F. (650°C.)	
5. Air-quench for handling	

The physical properties of such material will be

Ultimate tensile strength	70,000–80,000 p.s.i.
Elongation in 2 in	6–12 per cent

Another method of producing high-strength iron is to malleableize completely, modify the structure by redissolving the carbon, and then develop the matrix to produce the desired structure.

It must be noted that the heat-treatment of irons, as it has just been discussed, is a rather recent development and that we have been able only partially to discuss the many developments made and being made in this particular field. These special irons are finding wide application in places where formerly steel castings or steel forgings were required for the job. The heat-treated irons have sufficiently high tensile properties and shock resistance to be successfully used in such applications as machinery cams, connecting rods, conveyor chains, dies, gears, differential carriers, universal housings, etc. In addition to these properties, they are superior to steel in wearing qualities, stress-dampening properties, fatigue resistance, and machinability.

Suggested Questions for Study and Class Discussion

1. Describe the structural changes on cooling from above the point of solidification to room temperature and the structure at room temperature of the following irons: total carbon 2.50 per cent, silicon 1.00 per cent, manganese 0.25 per cent, phosphorus 0.15 per cent, sulfur 0.05 per cent; and iron of the analysis, total carbon 3.25 per cent, silicon 2.25 per cent, manganese 0.65 per cent, phosphorus 0.15 per cent, sulfur 0.10 per cent. Discuss the irons from a practical viewpoint.

2. What is the purpose of so-called *aging* of plain carbon cast irons?

3. How is it possible to obtain reactions in the heat-treatment of cast iron that are comparable to those in steels?

4. What effect does tempering have upon hardened cast iron, in regard to its strength and hardness properties?

5. Why is white cast iron seldom heat-treated by quenching and drawing?

6. Describe the typical "long-cycle" annealing process used in the production of malleable iron.

7. Describe the structural changes that take place from the original casting to the final product, using the long-cycle anneal.

8. Why is silicon of such importance in the production of malleable iron?

9. How may the long-cycle process be materially shortened? Discuss this from a structural standpoint.

10. Is it possible to produce malleable iron that is comparable to steel castings or steel forgings? Explain.

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