ESSENTIAL METALLURGY FOR ENGINEERS

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THE CAUSES AND THE CONTROL OF METALLIC PROPERTIES

BY

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WITH AN INTRODUCTORY PREFACE
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INTRODUCTORY PREFACE

By PROFESSOR R. S. HUTTON, M.A., D.SC.

ENGINEERING students have for long been awaiting some modern treatment of the subject of metallurgy, which would give them a better view of the possibilities in existence for the control of the mechanical properties of metals.

The author has made a bold attempt to supply this need and his treatment of the subject presents several novel and useful features.

It is noteworthy that as an Education Officer of the Royal Air Force he has come up against this urgent need of clarifying our views on mechanical properties and co-ordinating them with the causative sub-microscopic structures.

More than twenty years ago the late Professor C. F. Jenkin, who himself had been responsible in the last war for aircraft materials for the R.A.F., in a paper to the Institute of Metals entitled "On the Metallurgical Information required by Engineers," made an appeal for a new outlook. This book by Dr. A. C. Vivian is, I believe, the first suitable response.

The author has not attempted to cover too wide a field, but has realized that the engineer wishes to know what really controls the properties of metals and alloys. Amongst other outstanding features of the book he has given special attention to heat treatment processes and dealt with the subject in an original way, leading to a simplified outlook covering the many diverse processes. To many engineers heat treatments are still purely empirical workshop methods, whereas to-day both with steels and light alloys many of our vital structures demand delicate and exact thermal transformations, the nature of which must be understood to be controllable.

I feel sure that this book will serve a useful purpose in giving engineers and other students a stimulating interest in metallurgy and will assure that they approach their problems of the selection and utilization of metals by a much more scientific road than has hitherto been readily available.

CAMBRIDGE, 1942.

AUTHOR'S PREFACE

THE author has made an attempt to supply engineers, technicians and students with a short account conveying complete understanding of the essentials in modern control over the properties of metallic materials.

Before setting out to do this, he has had good opportunities of observing reactions to the more usual teaching methods, and has concluded that he should not be over ambitious in the matter of detail, nor too lengthy for sustained interest; but that his presentation should be extremely pointed and insistent on essentials, and themes of major value to engineers and technicians should survive any personal preference for unbiased metallurgy if the result is not to fall between the proverbial two stools.

The key to much of his particular view of the subject for the above purpose is to be found in a cautious adjustment of ideas about the mechanical properties on the one hand, and about the internal structures of metals on the other, until the two fit into and account for one another. In the belief that memory can be spurred by interest in analogies and comparisons far better than by trite classification or by the mere marshalling of facts, some trouble has been taken in the exhibition of kinships between different phases of the subject, and in showing the position of metallurgy within the whole realm of applied physics.

If, in so trying, the author seems to have presumed unduly upon the outcome of many profound researches still making steady progress, he pleads that the readers for whom this little book is intended are urgently calling for a clear snapshot of physical metallurgy as it is to-day: these readers could make but little use as yet of the fascinating though somewhat misty slow-motion films of the researches which advanced students of pure metallurgy expect.

The author's colleagues and friends have contributed valuable encouragement and direct assistance of various kinds. To all of them, particularly to those who have so helpfully and patiently borne with him in discussions of his own initiation, he tenders grateful thanks.

This Second Edition has derived benefit from criticisms, received with thanks, which have enabled the author to attend to certain ambiguities and errors in the First Edition.

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ESSENTIAL METALLURGY FOR ENGINEERS

CHAPTER I

INTRODUCTION

THERE appears to be one particularly direct approach to the very core of physical metallurgy which should commend itself to engineers under intensive training, on refresher courses, or anxious to keep abreast of the times. This is by way of deliberate answers to the questions—

> "By what internal mechanism are the metals naturally possessed of the mechanical properties—strength, ductility, and so forth; and how may metallic materials deliberately be endowed with all those properties to a high degree?"

Some method of presenting the subject consistently from the above angle could hardly be better adapted to an engineer's needs, and should be satisfying if it were to cover all modifications of the properties which might accidentally or intentionally be induced in metals by combinations of working, heating, or cooling. The unaided conception of "cohesion" has been so taxed in ill-starred attempts to explain these properties that it is high time a more convincing account was attempted.

Based on the space-lattice theory, and on a modified theory of amorphous cement, and involving no dangerously misleading simplifications, the story of arranging the metallic structures to provide the properties desired by engineers could be made to appeal to all of a practical turn of mind.

What does an Engineer want with Metallurgy?

The testing of materials comes into an engineer's training in the ordinary way; the *chemical* metallurgy of the production of metals from their ores is not essential; but it is this *physical* metallurgy which is all-important to him nowadays.

For things are not what they were; relentlessly, they tend to become more complicated for all—including the engineer! In these changing times few things change more rapidly than the metallic materials an engineer is given to fashion, to fit, and to put into use; nothing could be more rash than to suppose that such materials are robust in the sense that they can be shaped, finished, and built in without any idea of the subtleties which have been introduced into them by the metallurgist. It would probably be wiser on the whole for the engineer to suspect that his material might prove to be as useless as glass or as putty for his purpose, than for him to feel assured that any "ordinary" treatment according to common sense would be "good enough."

For, depending upon their own peculiar character and upon their previous treatment by heating and cooling, and by cold and hot work, different metallic materials of all kinds react in very different ways to further treatments of all kinds; so that the number of combinations of possible treatments which might be disastrous when carried out by the uninitiated is very great—enough, in fact, to warrant much attention on the part of engineers to the methods by which metallurgists endow metals with the necessary mechanical properties.

Unless an engineer approaches his material from the same angle as the metallurgists who produced it, he is pretty certain to be disgusted and defeated. Blindly to attempt to follow mere instructions is equally to invite disaster in far too great a number of cases. Seeing eye-to-eye in these matters with the producers is likely to make all the difference to engineers between success and failure.

Preliminaries to such a Short Course

The short cut referred to is through somewhat unaccustomed districts in which it should soon be possible to find the way about. It will be necessary to turn to elementary considerations of the structure of matter as ably demonstrated by chemists and physicists who have had the use of X-rays in their investigations. It will be desirable to adopt a rational and simplified outlook on the mechanical properties of metals, and to take a rather uncommon interest in the real essence or inner meaning of these properties, and in their relationships.

Metallic Structures

The meaning of "metallic structures" should be clear from the start: the phrase as used in this Course refers to the arrangement, disposition, or architecture of atoms within the metallic mass which is itself and directly responsible for the effects causing strength, ductility, etc., or for the lack of those effects giving rise to weakness, inductility, etc. Appearances to the naked eye, or under the microscope, may sometimes be described as "structures," but these never have much effect *per se* in modifying the properties; there is comparatively little in a microscopic view of a metal which can be interpreted usefully as causes of the properties; for these causes, for the mechanisms giving rise to the properties, it is necessary to go deeper into the invisible or atomic disposition, or "metallic structure."

CHAPTER II

THE AMORPHOUS STRUCTURE

IT is first necessary to explain the nature of the simplest case—the pure metal—and to explore the natural causes of the properties.

The Liquid or Amorphous Structure

The pure metals are "elements," or substances which have never yet been split up into component parts of different kinds. In the molten state they share with all liquids the property of fluidity, which is just the capacity of one portion of the mass to flow over another portion easily or with difficulty, but—an important point—with equal ease or difficulty *in any direction* whatsoever.

In all fluids the millions of tiniest particles (which we can call "atoms" for simplicity without doing any harm) are disposed quite irregularly throughout the mass; the mean distance between the atoms falls as the temperature falls. When hot, liquids are mobile; the flow of one portion over the other is relatively easy; but, as the temperature falls towards the freezing point, this flow is not so easy since the liquids become more viscous; the force of their own weight is not quite so effective in producing flow. An attempt to cut a cool viscous liquid will be resisted more than an attempt to cut a hot mobile liquid.

There is a force of attraction between all atoms which increases enormously as they close in on cooling; this force resists flow, section, or disruption more when the liquid is cold.

Now, all metals alter in structure (in the disposition of their atoms) at temperatures called their "freezing points," and this new highly important "crystalline" structure will be considered at length; but we have not yet sufficiently considered the irregular or "amorphous" structure peculiar to all fluids, including liquid metals and metals which are liquid-like even though they may be below their freezing points.

For a moment suppose it were possible to cool metals down below their freezing points, having entirely prevented any change of structure except this mere coming together of the atoms which always happens when substances cool. What would be the result? With metals such an experiment has not been successful hitherto, but with other substances like glass it has been possible to see what happens. There is a rapid increase in viscosity with fall in temperature until, at low enough temperatures, the extremely viscous liquid seems solid. This "solid" effect is no more than a high viscosity, for no change of structure has occurred between the *mobile liquid* glass at furnace temperatures and the *viscous liquid* glass at room temperatures.

Fused quartz is an example of a liquid which can, by quenching in water, have its amorphous structure trapped, caught, and retained at room temperatures: its properties can be fully examined and noted. This same fused quartz can also be allowed to change its structure normally at its freezing point—it can be allowed to "crystallize"—and the properties of this variety can also be examined. These latter are found to be different to those of the amorphous variety in many very important respects.

Wherever and from whatever cause the atoms of a metal at room (or common low) temperatures occur in a comparatively haphazard, random arrangement, there will be found certain mechanical properties peculiar to very viscous liquids. We should now consider what these are—

Pitch, another liquid, is rather easier to understand because, at ordinary temperatures, it has the kind of intermediate viscosity which enables us to appreciate the property change more easily. In reality, pitch at low temperatures would behave like glass or amorphous silica (fused quartz), whilst glass or amorphous silica would behave like pitch if their temperatures were raised.

In the case of all these amorphous structures their properties vary from those of mobile liquids to those of excessively viscous liquids (which pass as quite solid) depending entirely upon their temperatures.

The significant points about them are—

(*a*) They will exhibit some little elasticity and not a little strength if stresses are rapidly built up in them by fast loading; but, under these circumstances, they have no ductility at all—they snap like glass.

(*b*) On the other hand, they will flow like perfect liquids if the loading is so gentle that only the smallest stresses are induced in them for a very long time; in these circumstances they exhibit infinite ductility (flow like liquids), but no strength worth speaking of, and no elasticity, for the stresses have never attained the values which give these properties any significance. For example, a jagged lump of pitch placed in a wineglass would fill the vessel in time—exactly like a mobile liquid—under the force of its own weight.

To summarize: Amorphous materials have mechanical properties which vary widely according to (i) the temperature, (ii) the rate of loading. They vary as widely as possible in properties between the extreme limits of (*a*) high elasticity and strength with no ductility, (*b*) no elasticity or strength, but infinite ductility.

Metals always contain some, and often contain much, Amorphous-like Metal

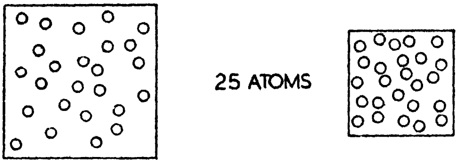
Up to the present it has not been found possible experimentally to trap amorphous metal by quenching fused metals; but this may be done in time, and in any event that does not in the least alter the fact that metal of the amorphous or similar structure occurs in metallic masses in everyday use for *two* very good reasons. Hence, all the foregoing about the amorphous structure is essential; metals in everyday use exhibit to some extent the curious properties above indicated because of the occurrence in them of some of this amorphous-like structure.

Grain-boundary Cement

It will be seen that, at their freezing points, liquid metals change most of their amorphous structure for a "crystalline" or beautifully arranged one; but these crystal "grains" are many, and though each is beautifully arranged inside they must mutually interfere at their outside surfaces in contact, giving rise to thin zones at which the atoms are not so regularly spaced but scattered in a relatively haphazard manner.

It does not seem to be a matter of essential importance whether the grain boundaries are zones in which the crystals suddenly change to a truly amorphous metal, zones in which the crystals become broken up into the tiniest unorientated crystallites, or zones in which an orientation of one kind is in some way strained round to adjust itself with an orientation of the neighbouring kind; for, in all the above suggestions, the usual crystalline regularity is lacking, and the properties peculiar to amorphous metal will assert themselves to some degree.

This is the first way in which metals come by an appreciable amount of an amorphous-like structure called the "boundary cement"; the boundaries, then, have some of the properties of pitch, tending to be elastic and strong in the cold, and where (as in ordinary testing) the stresses induced are medium high; but it has some capacity to flow and tends to be weak when the temperature is high or when the stresses are low but prolonged.



I. THE LIQUID (AMORPHOUS) STRUCTURE

THE METALLIC ATOMS IRREGULARLY DISTRIBUTED

AT HIGHER TEMPERATURES

THE MEAN DISTANCES OF THE ATOMS ARE GREATER, GIVING RISE TO EASIER FLOW OR SHEAR, WHICH IS MOBILITY IN A LIQUID AND HIGH DUCTILITY IN A SO-CALLED SOLID WITH THE AMORPHOUS STRUCTURE.

AT LOWER TEMPERATURES

THE MEAN DISTANCES OF THE ATOMS ARE SMALLER, GIVING RISE TO GREATER DIFFICULTY IN FLOW AND SHEAR. WHICH IS VISCOSITY IN A LIQUID AND LOW DUCTILITY IN A SO-CALLED SOLID WITH THE AMORPHOUS STRUCTURE.

N.B. 1. AS THE MEAN DISTANCE BETWEEN THE ATOMS IS REDUCED WITH FALL OF TEMPERATURE, THE ATTRACTION BETWEEN THE ATOMS (COHESION) RISES AT A VERY MUCH MORE RAPID RATE.

2. AT ANY ONE TEMPERATURE (THAT IS, FOR ANY PARTICULAR MEAN DISTANCE OF THE ATOM) THE PROPERTIES OF STRENGTH AND DUCTILITY WILL VARY WITH THE RATE OF LOADING. EVEN WATER WITH ITS MOBILITY WILL NOT FLOW (NOR PROVE DUCTILE) UNDER A FAST BLOW: EVEN VISCOUS (SO-CALLED SOLID) PITCH WILL FLOW (OR PROVE DUCTILE) IF URGED SUFFICIENTLY GENTLY.

FIG. 1

Evidence of the last two properties is forthcoming as follows—

(*a*) As the melting point of a solid metal is approached, the boundaries of the crystal grains become more like mobile liquids, and fracture of the metal is more readily produced by failure of the grain boundaries *instead of the usual* failure of the crystalline metal itself.

(*b*) Very prolonged tests at low loading (Creep Tests) exhibit a metal's high ductility owing to the greater flow in such circumstances of the boundary cement; the finer the grain, the more the boundary cement, and the greater the creep is found to be.

Ordinary metal, containing millions of grains per cubic inch, owes an important part of its mechanical properties to this "cement" without a doubt. This can be shown by applying a test like the Tensile Test to specimens of the same metal of greatly differing grain size, i.e. greatly differing amounts of boundary cement.

Single Crystals

It is possible by certain methods to make a test piece of metal consisting of one grain only, or of two or three grains only. On test it is found that the properties are extraordinary: elasticity is practically absent, and *at the outset* strength seems to be low while ductility is high. During tests of metals consisting of two or three grains it is evident that the boundaries are relatively strong regions which resist fracture and distortion in the tensile test better than the crystals themselves.

As the distortion of the test piece progresses towards the fracture of these specimens, the properties rapidly change as an amorphous-like metal of a different origin appears, which is sometimes called "the amorphous material of cold work."

Grain Refining

The above suggests the kind of effect to be expected if very gross coarsening of grain in ordinary metal were to take place; but, in common practice, the differences in grain-size produced by technical refining, though great, are usually within the range where gross changes in strength and ductility do not take place. It is noticeable that, in refining a metal's grain, strength and ductility are both somewhat improved as a rule, the extremely tenuous boundary cement of the usual amount exhibiting some of the strength and some of the ductility of an amorphous material with properties intermediate between the extremes possible.

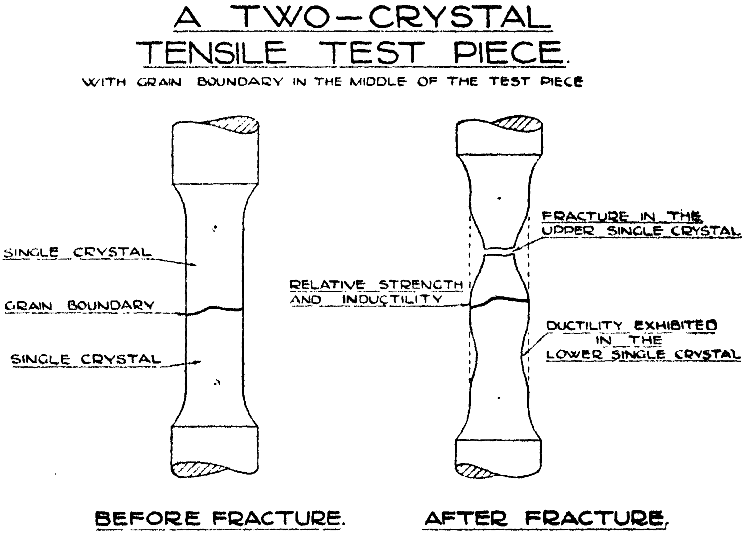
At somewhat unusual extremes of grain fineness, strength and hardness rise with astonishing rapidity as grains get smaller, and the predominating influence of these boundaries asserts itself.

" Shock " or " Impact "

The outstanding feature of a fine-grained metal is its resistance to the *development of a crack already started*—measured by the energy absorbed in the so-called "shock" or "impact" test; it is chiefly on this account that practical grain-refining is carried out, and this point will be considered later as it does not depend on the nature of the boundary cement so much as on the more frequent change in the direction of the fracture path through the crystals on account of the increased number of grains.

Fatigue

The fractures under fatigue conditions, or in Fatigue Testing (to be considered later), are noticeably on the grain boundaries and "amorphous" slip bands until so much of the specimen's effective area of cross-section has been ruined by this means that the remainder fractures in the usual way—mainly across the residual crystalline material of the grains themselves. In fatigue tests, the reversals of stress at a high rate may yield effects absent from familiar static tests; but a more significant feature still is the effect as "stress raisers" of tiny cracks or flaws in the metal or at its surface; these enormously *increase the local loading rate* so that it is not surprising in these circumstances to find that the "amorphous" metal of the boundaries and of cold work ruptures first; nor is it surprising to find that hair cracks (the most potent stress raisers) are just the blemishes in metals which make for failure under fatigue conditions.



PROVING THE RELATIVE STRENGTH AND INDUCTILITY OF THE METAL AT THE GRAIN BOUNDARY; FRACTURE TAKES PLACE EITHER ABOVE OR BELOW THE BOUNDARY INSIDE A CRYSTAL BUT NOT AT THE BOUNDARY. LIKEWISE, POLYCRYSTALLINE MASSES OF METAL ARE PLASTICLY DEFORMED BY THE DUCTILITY OF THE GRAINS; AND THE FRACTURE-PATH TRAVERSES THE GRAINS THEMSELVES, NOT THE GRAIN BOUNDARIES (EXCEPT IN FATIGUE, AND POSSIBLY IN NOTCH TOUGHNESS TESTS.)

FIG. 2

The "Amorphous Material of Cold Work"

This is the second way in which metals come by "amorphous" structure—inside the crystal grains themselves this time. It is formed, no matter if the metal is fine or coarse grained, as a result of stressing the metal beyond its elastic limit (its "primitive" or original elastic limit) until it is permanently deformed in shape; the greater the distortion, the more does the "amorphous" metal appear, and the stronger and less ductile does the resulting metal become under the medium fast loading conditions in such a test as the tensile test. Metal strained to the point of fracture is as fully charged with "amorphous" metal as it ever will be; and, though strong and inductile towards loading rates which are common in tests, they are comparatively weak and ductile under the loadings of such tests as the Creep Test; in other words, this "amorphous" metal is as much like pitch as is the boundary cement.

The "amorphous" metal of plastic distortion which has been carried out at temperatures below the "annealing" temperatures, or the "amorphous material of cold work," forms inside the crystal grains; and until the nature of crystalline structure has been fully discussed, the cause of its formation cannot be so well explained. It should be sufficient for the time to mention that slip or glide along certain sets of planes inside the grains (together with a flow of adjustment in the grain boundaries) is responsible for all the change of shape which a metal undergoes when it is permanently distorted in the cold, or at common temperatures. During this slip, some of the original crystalline metal is broken up or jumbled either into an excessively fine state of tiniest crystals, or into a truly amorphous state, in thin zones wherever motion to produce the distortion has actually taken place inside the grains.

There is some X-ray evidence to show that cold work has no power to reduce crystallite size below about one hundred thousandth of a centimetre; but even this reduction into completely unorientated crystal fragments is a considerable step towards the production of a truly amorphous state in metals. A little more of this rather less amorphous metal would be as good as a little less of a truly amorphous metal as far as properties are concerned, and in any event nobody has discovered very much about the actual *quantities* of this metal obtained by cold work for any degree of cold work, except that metal on the point of fracture seems to consist *entirely* of unorientated crystallites of 10^{-4} to 10^{-5} cm. size.

Properties affected by "Amorphous" or "Near-amorphous" Metal

The mechanical properties are determined by tests—strength and ductility by the tensile test; such tests are relatively fast for materials like "amorphous" metal, registering to some degree the elasticity, strength, and inductility of that kind of metal at ordinary temperatures.

It cannot be too well remembered, however, that such metal will behave in the opposite sense, and will exhibit negligible elastic limit, weakness, together with ductility if ever it is tested by prolonged low loading, or at elevated temperatures. Naturally it will also exhibit intermediate properties if tested at medium rates of loading or at medium temperatures, because amorphous materials of all sorts have properties which gradually vary from one extreme to the other as loading rates and temperatures vary between the extremes.

Ordinary metal, then, always contains some "amorphous" metal as boundary cement, and frequently contains much more as a result of plastic distortion in the cold; it will therefore tend to have mechanical properties varying (according to the actual amount present) with the speed of testing and with the temperature of testing.

Consistent Test Results

As the common speed of testing and the common temperature of testing are fairly constant, and as the amount of boundary cement in the usual specimen for test does not vary very much, we can expect consistent test results for any one metallic material. Large changes in any one of these factors will become evident in inconsistent results registering differences in mechanical properties.

The Old Amorphous Theory and its Modification

The penultimate paragraphs on pages 6 and 10 will be found to distinguish between the truly amorphous structure of the liquid (mobile, viscous, or so viscous as to be rigid) on the one hand, and the near-amorphous, amorphous-like or "amorphous" structures of the grain-boundaries and of the slip-bands on the other hand.

None of the distinctions urged in more recent years between the truly amorphous structure of the liquid and the arrangement of the atoms at grain-boundaries and slip-bands has vitiated an argument that crystalline structure has at those places *degenerated in the direction of* an amorphous structure; and that, to the extent of the degeneration, the crystalline properties have therefore been modified to approach the properties of a truly amorphous matter at those places. To that extent an Amorphous Theory must still hold good; with the difference that the application of the theory in detail is now better understood.

CHAPTER III

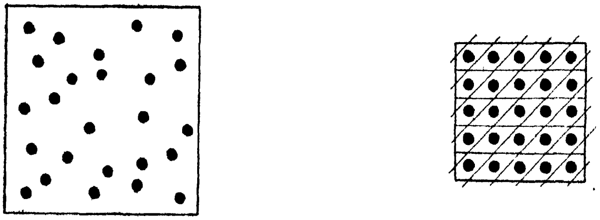
THE CRYSTALLINE STRUCTURE

Crystallization

THE atoms or tiniest particles of all substances are in a state of agitation by heat, but they vibrate and rotate about *mean* positions which can be assumed to be fixed when considering structures.

CRYSTALLISATION.

THE CHANGE FROM THE AMORPHOUS STRUCTURE TO THE CRYSTALLINE STRUCTURE AT THE FIXED TEMPERATURE OF THE FREEZING POINT.



THIS CHANGE (AT ONE TEMPERATURE) INVOLVES:-

1.- A SUDDEN REDUCTION OF MEAN DISTANCE → 25 ATOMS AMORPHOUS, OCCUPY MORE SPACE THAN 25 ATOMS CRYSTALLINE

2.- AN ORDERING OF THE ATOMIC ARRANGEMENT IN A REGULAR PATTERN OR "SPACE LATTICE.

IN CONSEQUENCE OF (1) ABOVE, THE COHESIVE FORCES BECOME SUDDENLY GREATER; AND IN CONSEQUENCE OF (2) ABOVE, THERE APPEAR PREFERENTIAL PLANES OF FLOW OR SHEAR, WHICH WERE NON-EXISTENT IN THE AMORPHOUS STATE. BOTH OF THESE CAUSE THE GREATER RIGIDITY OF THE CRYSTALLINE AS COMPARED WITH THE LIQUID (AMORPHOUS) BOTH AT FREEZING TEMPERATURE; BUT THE MECHANICAL PROPERTIES AT THIS HIGH TEMPERATURE, ARE VERY POOR, AND WILL CONTINUE TO BE SO, UNTIL AT MUCH LOWER TEMPERATURES, THE LATTICE SHRINKAGE ENORMOUSLY INCREASES THE COHESIVE FORCES.

FIG. 3

When liquid pure metals are cooled, they solidify at fixed temperatures, undergoing the change which is known as "crystallization." The essence of crystallization is change of structure giving rise to change of properties; for the structure of any liquid consists of atoms in a haphazard arrangement called an "amorphous structure" permitting the passage in any direction of one atom past its neighbours, this bestowing the property of fluidity; whilst the solid called "crystalline" has a structure of a patterned or perfectly regular type, the atoms of which are bound by some degree of force to remain fixed in the pattern, this arrangement bestowing the property of (crystalline) rigidity, etc.

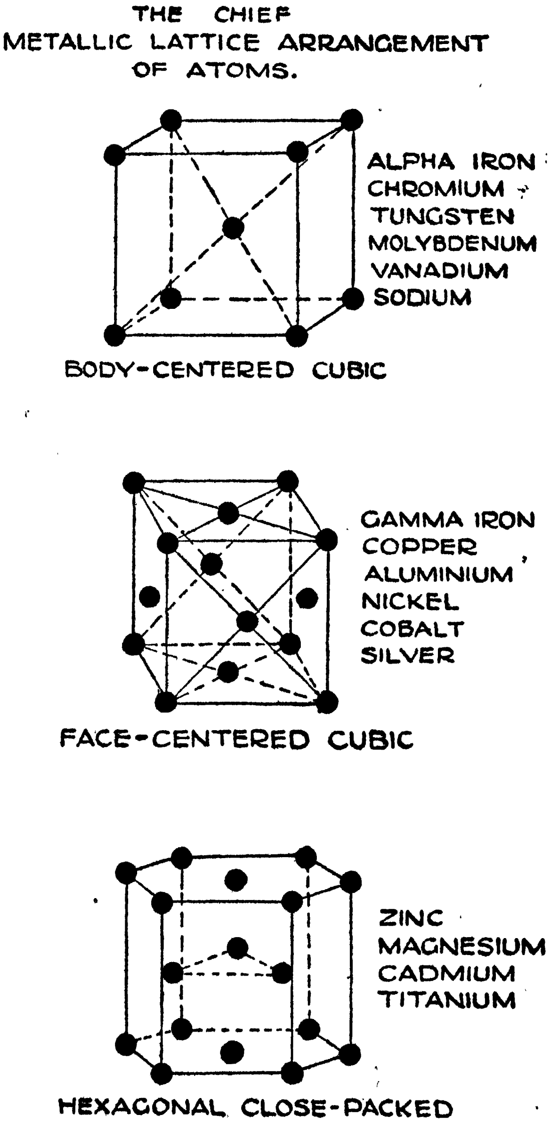


Fig. 4

A "centre of crystallization" is a point in the liquid from which the regular pattern or "space lattice" grows. If it were possible to restrict the centres of crystallization to one centre throughout the whole operation of freezing a metal, the resulting solid mass would consist of a *single crystal*: it is difficult to do this, and in ordinary practice quite impossible. As a general rule, freezing in a metal mass takes place from a very large number of centres all at the same time, and crystals extend from centres until they meet and interfere with crystals from neighbouring centres, producing irregularly shaped grains each of which is a perfect crystal inside, though lacking crystalline external form on account of this mutual interference.

Single crystals and *polycrystalline masses* differ in properties because of the difference in structure: the grain boundary cement of the latter constitutes a structure entirely absent in the former, and mixes its peculiar mechanical properties with those of the crystalline structure.

Different pure metals crystallize in "lattices" of different types, the commonest being the cubic type—either the "body-centred" cubic, or the "face-centred" cubic type. The atoms of the crystalline metal are usually called the "units" of the lattice.

The Rigidity of the Crystalline Structure

By comparison with the ordered structure, the random disposition of the atoms, as in a liquid, seems to permit of an easier flow of one part of the mass over another; this is because the flow or shear motion of a liquid is equally easy or difficult in any direction whatsoever, so that the number of possible and available shear planes is infinitely great.

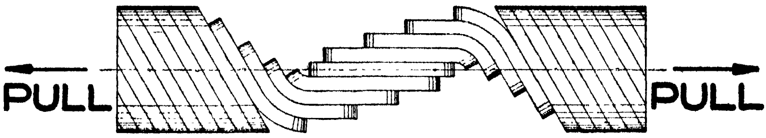
In a liquid, whether it be mobile or viscous, there are no planes of *preferential* flow: but in a crystal the lattice is such that there are preferential planes of flow and the number of available shear planes is limited. The causes, then, of the relative rigidity of the crystalline structure as compared with liquid structure at the same (freezing) temperature are—

(i) There is a certain sudden closing of the distances between atoms on crystallization.

(ii) The number of flow or easy shear planes is considerably reduced in the mass.

Gliding Planes

These latter are called the "gliding planes" of a crystal; when any force from outside makes the crystal exert more than a certain limiting stress in opposition to it (the stress of the elastic limit), plastic change of shape or flow takes place by the motion of one part of the crystal over another at these gliding planes; the displacement is rendered visible under the microscope by the development of "slip bands" in the crystal grains. At the very outset, motion of this nature is comparatively free to take place; but opposition rapidly mounts as motion proceeds, rather as opposition to the forward motion of a derailed train would mount as the track became increasingly destroyed by such motion.



A "SINGLE CRYSTAL" TENSILE TEST PIECE.

EXHIBITING PLASTIC DISTORTION BY MOTION AT THE LATTICE PREFERENTIAL PLANES OF FLOW OR SHEAR – THE LATTICE GLIDING PLANES.

SIMPLEST DIAGRAMMATIC ILLUSTRATION OF MOTION ON ONE PLANE ONLY, WHEREAS MANY SETS OF PLANES ARE REALLY BROUGHT INTO ACTION DURING THIS DISTORTION.

FIG. 5

The original crystalline or orderly arrangement of the atoms becomes destroyed in the region of the planes, giving rise to zones of metal having a structure more like that of liquids. This "amorphous" metal is therefore much as liquid metal would be at temperatures far below the melting point: for the sake of simplicity it may be regarded as undercooled liquid metal, having to some extent the peculiar properties of pitch or glass—properties possessed by liquids with exceedingly high viscosity.

Change of shape of the usual Polycrystalline Mass

Change of shape in the usual metallic mass is effected by the combined mutually adapting change of shape of the constituent crystal grains; these grains, besides being individually sheared along the various sets of gliding planes which come into operation, tend to be orientated so that any long axes or easy gliding planes they have are turned in the direction of the motion produced by the load; they also tend to be lengthened in the direction of motion, and shortened in the direction at right angles: they are thus *drawn out* in the direction of metal "flow," producing a visible crystalline "flow structure"; so that irregular polygonal grains very soon look rather like packed sardines, and ultimately like parallel-running fibres.

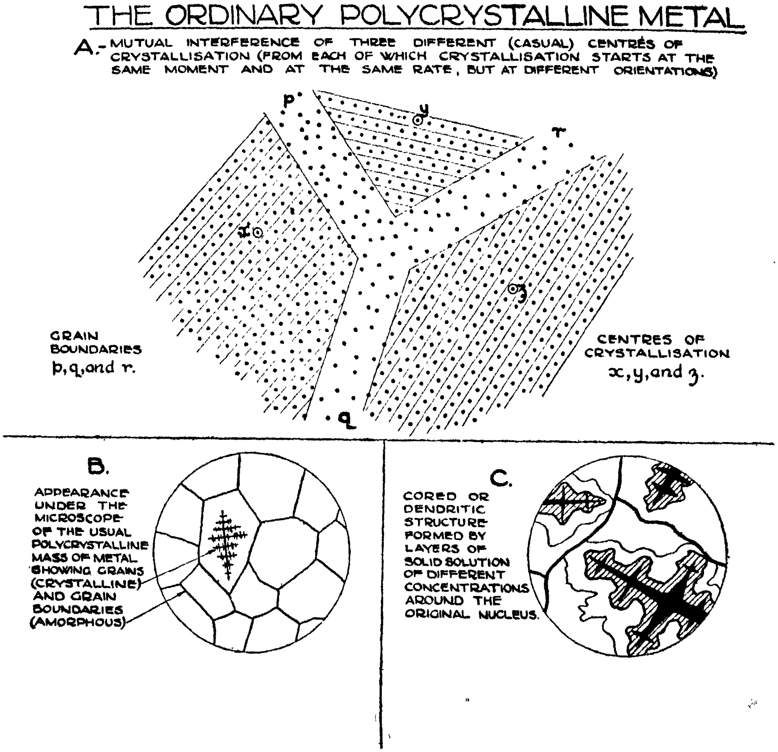


Fig. 6

Strengthening by Cold Work

Such crystalline masses consist of lengthened grains all of which contain innumerable layers (in many directions throughout) of "amorphous" metal. A new structure has been obtained by cold work, the new property of added strength has been acquired by cold work, with loss of ductility which may or may not be a serious consideration for the object in view.

At elevated temperatures, different for different metals and alloys, this "amorphous" metal produced by cold work is changed back again to the common crystalline metal; these are called the "*recrystallizing*" or "*annealing*" temperatures; there is no disappearance of the boundary cement, however, though long exposure to high temperature does coarsen the grain. The recrystallization from a great number of points in the amorphous metal has the effect of *reducing the grain-size.*

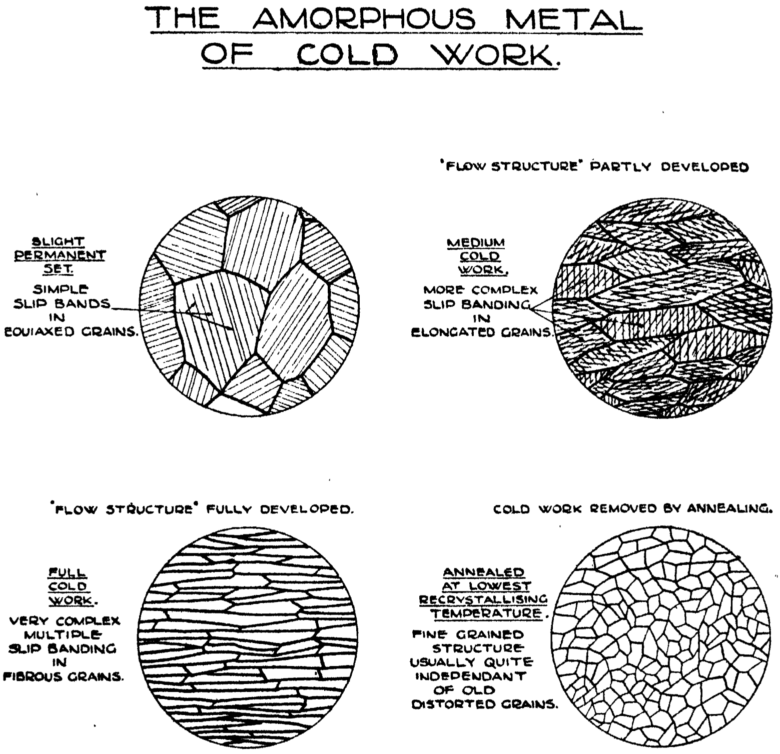


FIG. 7

Hot Work

Hot work done on metals at or above the annealing temperatures permits plastic change of shape to take place reasonably fast to any desired degree without loss of the practically infinite ductility of metals at these temperatures and distortion rates. If it is carried on down to a sufficiently low temperature, but not too low, grain-refining will result.

In a preliminary and superficial manner these chapters have gone actually beyond answering the first part of the question in the Introduction. It will be noticed that, from the paragraph "Gliding Planes," a start has been made in reply to the second part of the question—"How may metallic materials deliberately be endowed with all these properties to a high degree?" For metals are not *naturally* cold worked; and cold work on metals is a recognized *artificial* means of inducing certain desired properties in cast metals, copper, zinc, brass, etc., etc.

Together with the *strength* induced by the presence in cold-worked metal of "amorphous" metal, there are also extra *elasticity* and *hardness*. These three improvements in properties may be total gains if not induced at serious sacrifice of ductility. Cold work is a valuable method of modifying properties.

CHAPTER IV

SOLID SOLUTIONS

The Structures and Properties of Alloys

THE pure metals differ among themselves in properties according to differences in lattice type and size, or in the size of the units, or in the forces exerted by the units. In any one pure metal, modification of grain-size modifies properties to some degree, as also does cold work.

But another important and methodical control of mechanical properties may be imposed upon a pure metal by mixing or "alloying" a different pure metal or element with it.

As a general rule (to which there are exceptions like copper-lead), metals are soluble in one another to an unlimited degree when molten; in some cases solubility persists to the fullest degree in the solid state as well. Copper-nickel or copper-manganese alloys of any composition consist when solid of one kind of crystal only, a crystal with a lattice of the paramount metal harbouring units of a foreign metal until, with sufficient quantity of the latter, it could be described as a mean between the true lattices of the component metals. Such alloys are called "solid solutions," and their crystals "mixed crystals," because the one lattice contains both metals.

As a rule, units of the original metal are *replaced* by units of the foreign metal.

In the majority of cases, however, even where there has been a complete range of liquid solubility there is only a limited range of solid solubility; silver and copper, for example, although mutually soluble in all proportions when molten, only form solid solutions in the ranges from 0 per cent to 6 per cent of copper in silver, and from 0 per cent to 15 per cent of silver in copper (which is the same as from 85 per cent to 100 per cent of copper).

The Solid Solution

The curious lattice of the solid solution is a new structure and its effect on properties is far reaching.

The general effect of the foreign units is to render the true planes of the original metal less true by producing humps and hollows in them, thus "*pegging*" them and adding to the resistance to motion on them, and making the alloy stronger; motion on planes such as these results in more widespread damage to an ill-shapen lattice, giving rise to excessive "amorphous" material for very little motion, so that fracture takes place as a rule before the usual ductility of the original metal is displayed.

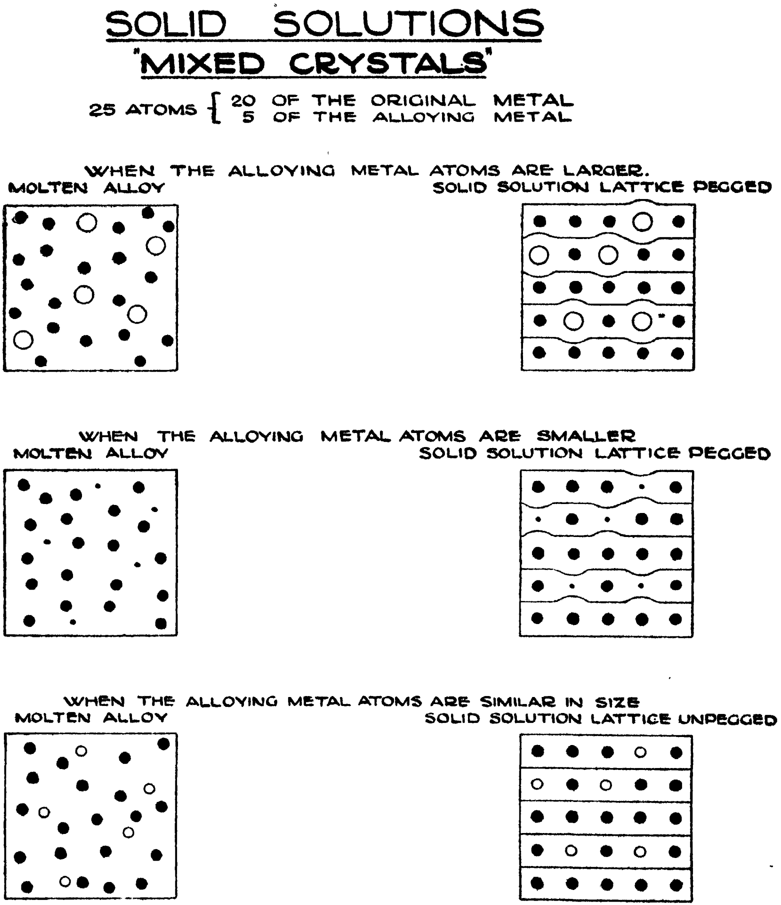


Fig. 8

The irregular disposition of foreign units in the lattice has been found to produce extra strength without much falling off in ductility.

The *regular* disposition of foreign units produces what is known as a "super-lattice" throughout the crystal, a structure giving good strength with a greater falling off in ductility. There may be an ill-defined compound present in those cases.

Phases of Alloys from Solid Solutions to Compounds

The amount of lattice change and property change, for ever-increasing proportions of the added metal, may *steadily increase for a phase*; then it may *increase at a greater rate for a second phase*; and, yet again, may *increase at a still greater rate for a third phase*; (these "phases" being composition ranges, which are sometimes called the alpha, beta and gamma phases or "constituents" and which are visible as distinct constituents of the alloys under the microscope).

The reasons for this effect is that certain groups of single lattices in the paramount metal (say groups of eight) permit of the replacement of certain of their units by the alloying atoms until the addition of the alloying metal has been sufficient to replace all of these *particular* units; the amount of distortion and property changes is directly proportional to the amount of the alloying metal up to the saturation of this phase. After this, the addition of extra alloying metal has to replace *certain differently placed* units of the groups of lattices, and the property changes are of a different order: the effect of replacing these units is often more drastic than the effect of replacing the first set; in this phase ("beta") the solid solution may have weak properties of a metallic compound which is capable of dissolving the component metals. After the saturation of this second phase by the replacement of the second set of units, there may be further replacements possible; but if so, the product obtained is usually a definite *metallic compound*, not a solid solution.

But it must not be supposed that every alloy exhibits such a complication of phases; most "binary" (two metal) alloys are simple, involving the alpha and beta phases only, where the alpha is the solution of the foreign metal in the original metal, and beta the solution of the original in the foreign. Where there is no solid solubility of the metals, the pure metal solid phase itself is called the alpha phase, or the beta—as the case may be, depending upon the side of the alloying range to which reference is made.

The development within the lattices of anything of the size and nature of a *metallic compound molecule* (group of atoms) seems to have a profound lattice "pegging" effect which, in some cases, brings with it a high degree of strength and inductility; such a *solid solution of a compound* may be brought back to a condition of adequate ductility with very good strength by some treatment to flocculate the compound to the desired degree, so that less of it is left in the lattice to peg it, the remainder being so coarse as to have no lattice influence. In other cases, good strength with good ductility is obtainable directly with ease.

All the above assumes that the foreign metal is having increasing effect in modifying an original lattice, which becomes unrecognizable as such when the development of the lattice of a compound takes place, or when the foreign metal reaches such a concentration that the predominating lattice is that of the foreign metal itself. As soon as the latter takes place, the effect of adding foreign metal becomes the equivalent of removing the original metal from the lattice of the foreign metal, and the matter is more easily studied from the opposite side; the gamma, beta and delta from the first standpoint become the alpha, beta and gamma from the second standpoint.

(The Greek letters alpha, beta and gamma are not always employed to refer to these phases: in the study of steels they refer to the three forms of pure iron, though "alpha" iron is sometimes meant to imply some solid solution such as nickel-iron, and "gamma" is often meant to imply some solid solution of carbon, nickel, etc., in gamma iron.)

CHAPTER V

METALLIC COMPOUNDS

What is a "Compound"?

MUCH of this presentation of metallurgy calls for a very clear understanding of the differences between a *mixture*, a *solution*, and a *compound*, of two or more components.

In the cases of the first two the differences are only of degree, a mixture being a coarse aggregate of the components, and a solution being an aggregate so fine as to be an intermingling of the very smallest conceivable particles of the components. These two states merge in the "colloidal solution" or emulsion, in which both mixture and solution can be said to occur together. (See Chapter VI.)

But a compound is entirely different: it is no aggregate derived by mere intermingling; it is an entirely new and different substance obtained by the *union of atoms of the components.* Heat is evolved when this "chemical" union takes place, and the result is a substance *unlike* any of its components.

The essential point about a compound, however, is that the smallest particle of it is a *bulky group of atoms coupled together* which constitutes a "peg" of a *size larger than that atomic peg* in the usual solid solution.

Whether the alloy be of one metal with another, or with a "metalloid" like silicon, or non-metal like carbon, there is a tendency towards the formation of a *compound* whenever the proportions of the components by weight are in the vicinity of *very simple atomic* proportions.

If one component has an atomic weight x, and the other an atomic weight y, there are tendencies to the formation of compounds at concentrations such as—

$3x$ with $1y$
$2x$ with $1y$
$3x$ with $2y$
$1x$ with $1y$
$2x$ with $3y$
$1x$ with $2y$
$1x$ with $3y$

In some cases the tendency is absent or very weak because the "valencies" (habitual combining capacities) of the metals are not satisfied by all of these combinations; or it may be strong enough to give rise to definite compounds such as $CuAl_2$, Mg_2Si, Fe_3C.

Now, compounds being completely different and perfectly definite substances, wherever they are formed they break the sequence of useful alloys between 100 per cent A and 100 per cent B, and lead to alloys between 100 per cent A and 100 per cent *AB compound* and between 100 per cent *AB compound* and 100 per cent B, in which the AB compound and its strong solutions in either component are commonly worthless, being far too weak and quite inductile.

COMPOUND IN MOLECULAR DISPERSION.

AT SOME TEMPERATURE AFTER CRYSTALLISATION, CHEMICAL COMBINATION BETWEEN THE ATOMS OF A SOLID SOLUTION GIVES RISE TO BULKY MOLECULES OF A METALLIC COMPOUND WITH ENORMOUS PEGGING EFFECT ON THE ORIGINAL METAL LATTICE

SLIGHT PEGGING EFFECT ON IRON LATTICE, CAUSED BY SLIGHTLY DISSIMILAR ATOMS OF CARBON IN SOLID SOLUTION.

RELATIVELY GREAT PEGGING EFFECT ON THE IRON LATTICE AS SOON AS MOLECULES OF Fe_3C HAVE FORMED BY CHEMICAL COMBINATION IN THE IRON LATTICE

FIG. 9

This is not to say, however, that these definite compounds do not play *an essential part* in modifying the properties of the lattices in which they are developed to a moderate extent. When the compound molecules are developed within the dominant lattice to a small degree, they have a profound pegging effect on a pure metal or solid solution giving strength—often without so much loss of ductility—which can be varied by permitting the compound molecules to flocculate (undergo segregation) to any desired extent, whereby they are removed from the lattice.

Thus, duralumin "age hardens" through the slow development of the molecules of Cu Al_2 within the lattice of aluminium; the martensite of steel is excessively hard through the development in the alpha iron lattice of the Iron Carbide molecules. Annealing heat treatment permits these to be reduced by flocculation. Any degree of flocculation, collection, or segregation of the compound molecules will reduce the strengthening effect, producing a "tempering" of the prior state. Ultimately, an entire disappearance of prior fully pegged properties will be contemporaneous with the full flocculation of the compound to coarse crystals easily seen under the microscope.

A serious warning would not be out of place here: there should be complete repudiation of any accidentally conceived idea that the properties of the lattice pegged strongly by a compound bear any relation whatever to the specific properties of the compound itself when in the massive state. For example, we are not in the least concerned with the strength, hardness, etc., of $Cu\ Al_2$, but only with the *size of the molecule for pegging* purposes in the aluminium lattice. For further example: martensite owes nothing of its great strength and hardness to the specific hardness of massive iron carbide, but only to the size of the carbide molecule as a peg in the alpha iron lattice.

Precipitation

Many have had experience in chemistry of producing "precipitates" from aqueous solutions, and have observed the coarse floccules of solids which occur. The word "precipitation" is therefore closely associated in the minds of many with the act of formation *plus* flocculation.

This "little knowledge" may be dangerous for the proper understanding of structures resulting from precipitation in alloys; a more complete conception is vital—

Stage I. "Precipitation" is a word often used in metallurgy to imply nothing more than the *formation in molecular dispersion*—the first stage.

In chemistry when sulphate of lead, for example, is first formed or produced by the addition of a sulphate solution to a lead salt solution, it is formed as *individual molecules of* $Pb\ SO_4$, or in molecular dispersion (or distribution) throughout the solution. Ultimately, as we know from observation, the $Pb\ SO_4$ floccules appear and sink to the bottom of the vessel; these floccules are aggregates or bundles of $Pb\ SO_4$ *crystals*.

Stage II. The eye cannot see the formation of $Pb\ SO_4$ in molecular dispersion, but only in the next stage when the molecules have come together (perhaps thousands of them) to form spherical agglomerations—liquid globules of very small size—giving rise to what is known as a "colloidal suspension" or a "*colloidal dispersion*," coarser than a molecular dispersion and visible as an opalescence throughout the solution.

Neither the molecular nor the colloidal state resembles the crystalline state in the least.

STAGE III. It is in such minute globules, however, that crystals form: the colloidal state is intermediate between the first "precipitation" and the crystal formation.

STAGE IV. The final stage has not yet arrived; when it arrives it is either chiefly crystal *growth*, or chiefly a *flocculation together* of tiny original crystals; or else there may be both growth of crystals and flocculation of crystals.

Precipitates commonly described as "crystalline" are visibly crystalline because there has been much crystal growth and comparatively little flocculation. Precipitates commonly described as "amorphous" are those whose crystals have remained tiny, and flocculation has masked the crystalline character; sometimes the microscope will show they are crystalline.

Precipitates commonly described as "colloidal" or "gelatinous" are those in which crystals have never formed at all, but the original liquid globules (above described) have flocculated together instead. (Certain physical conditions of precipitation favour the formation of crystals, others the formation of colloids, and many precipitates can be obtained in either form by varying conditions.)

Metallurgical Precipitation

Metallurgy is full of examples of the "precipitation" of compounds of the metals from liquid and solid solutions, but those precipitates from solid solutions are highly important because the properties of alloys depend upon them, and particularly upon the *form they take*: heat treatment is devised to control and determine the form they take.

The first "precipitation"—of molecular dispersion—from a solid solution is that form which will have the most effect in modifying mechanical properties by pegging the lattice and, in effect, producing humps and hollows on the otherwise plane gliding planes.

The smallest possible particle of a compound is its "molecule" which, being a group of atoms, is bulkier than a single atom; the pegging effect of molecules on the lattice is very great, but excellent ductility is not ruled out with such pegs, and the very best properties may often be induced by such pegs when present in the right numbers.

Every further step towards the formation of a *visible* compound is a step removing the lattice pegging molecules and, by coalescing them, preventing their stiffening effect on the metallic lattice.

The Metastable and Labile States

The formation of metallic compounds from liquid or solid solutions of metals as a result of cooling is similar to the formation of any other constituent, or of a precipitate in aqueous solution, inasmuch as that metastable and labile states may exist in all cases.

Whenever, according to strict equilibrium, a compound or other constituent is due to be formed, it will be formed only if there is a nucleus to form on. If a nucleus of some of the new phase is provided at or just below the equilibrium temperature of formation, the phase will be precipitated around that nucleus steadily but without haste, in fact rather slowly. On providing another nucleus, the speed of precipitation will be doubled because there are two nuclei. On providing a hundred nuclei, the speed will be a hundred times as great.

This state of affairs will continue unchanged for a certain range of temperature below the equilibrium temperature. This range is known as the *metastable range* in which precipitation is slow, and in which all the precipitation can be upon one nucleus to form one single massive crystal if desired.

Common or ordinary cooling provides no nuclei, of course: in their absence, no precipitation will take place in this metastable range; so that ordinary cooling continues *below* the metastable range into the *labile* range.

Upon reaching this range, the slightest shake, vibration, or commotion will start the compound formation *at a very large number of points all at the same time.* This is what is called the "labile shower" of the new phase, and it is of molecular dispersion. Such a shower or sudden formation will give rise *to evolution of heat* usually sufficient to bring the temperature back to the equilibrium temperature, or at least into the metastable range.

Thousands of nuclei have now been provided naturally, and precipitation continues steadily upon the nuclei supplied by the labile shower.

Consider the case of cooling considerably under the metastable range—undercooling which can be done if the solutions are not in any way disturbed. It must be remembered that, as temperature falls, viscosity or crystalline rigidity increases and will tend to hold up the migration within the solution (liquid or solid) which is called for in the *complete* precipitation of the new phase. At some stage of undercooling the labile shower might take place, but by that time the viscosity might be so great as to inhibit nucleal growth entirely or to slow it up so much as to render it negligible for all practical purposes.

If this happens, then a molecular dispersion, say, of a compound might be practically fixed and the mechanical properties conveyed by such a dispersion also fixed.

This may be an important consideration in the heat treatment of alloys.

CHAPTER VI

EUTECTICS AND EUTECTOIDS

IT has been seen that alloys must vary in properties as the lattice structure of the original metal becomes stiffened (gliding planes pegged) by the addition of the alloying metal (or element), and that such stiffening effect may not be smoothly and continuously increased as additions of alloying metal are made, but is liable to variation in intensity as new phases develop.

It has also been noted that the earliest stages in the development of a phase which is an actual compound has the most potent effect of all in raising the strength and in reducing the ductility, but that excellent compromises in these properties may be obtained by certain means.

In many alloys we are practically concerned with the structures of the above phases only; but, in the majority of alloys, an entirely different and a highly important constituent called the "Eutectic" (or "Eutectoid") makes its appearance and affects the properties to their detriment as a rule, but sometimes favourably.

The Eutectic

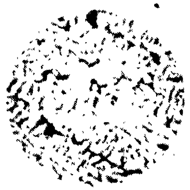
If two neighbouring phases—pure metals, solid solutions, or compounds—are not contiguous, and do not cover the whole range of composition between them, then into the range not so covered will appear a new constituent—a mixture (not a solution, but rather like a "colloidal solution") of the two neighbouring phases. In this colloidal form, which is a kind of emulsion, the phases are described as "secondary," whereas in the state *unmixed* they are described as "primary."

For example: Two metals which do not form solid solutions with one another at all, such as lead and antimony, have a 100 per cent gap in composition between them, and all these alloys will be "eutectiferous" (containing this eutectic constituent to a smaller or greater degree). Lead rich alloys will contain primary lead and the eutectic mixture of secondary lead and secondary antimony. Antimony rich alloys will contain primary antimony and the same eutectic as before.

At some fixed intermediate composition the whole alloy will consist of Eutectic: this is called the "Eutectic Alloy."

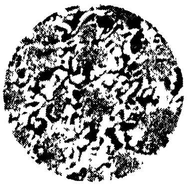
Alloys of metals which form a continuous sequence of solid solutions over the whole range show no eutectic; but, if in an alloy of A and B, B dissolves in A up to 10 per cent of B in A, and A dissolves in B up to, say, 20 per cent of A in B, there is a 70 per cent gap of compositions in which there are no solid solutions possible, and the eutectic constituent will appear in alloys between 10 per cent of B and 80 per cent of B. The eutectic in this case will be a fine mixture of maximum solid solution of B in A and of maximum solid solution of A in B (i.e. 10 per cent B in A and 20 per cent A in B). Again there will be an alloy of intermediate composition which has no primary phase, but consists of the eutectic only.

TYPICAL EUTECTICS

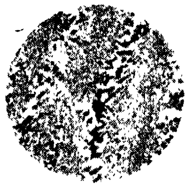


IRON - CEMENTITE

LEAD-TIN

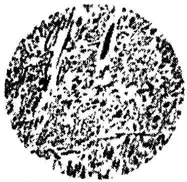


COPPER-COPPER PHOSPHIDE



ANTIMONY-ANTIMONY COPPER COMPOUND

TYPICAL EUTECTOIDS



$\alpha + \delta$ EUTECTOID (COPPER - TIN)

α + CEMENTITE EUTECTOID (STEEL.)

FIG. 10

Eutectic constituents are very fine crystalline mixtures visible as matrices in the microscopic views. These are in grains which are not crystals—only mixed aggregates of very fine "crystallites", *having a very different order of size* to the size of the crystal grain.

The appearance is characteristic and striking in many cases; in some cases it is very like that of a colloidal mixture or emulsion.

The Eutectoid

Whatever can happen during the freezing of a liquid solution to a solid alloy can also happen when, on cooling, a solid solution breaks into component parts. A constituent entirely similar to the one above described may come about from the same causes by the breaking down of a solid solution; when it does it is called the "Eutectoid" constituent.

(When concerned with a Eutectoid formation, the terms "primary" and "secondary" are not commonly employed; but instead, the words "free" or "pro-eutectoid" correspond to "primary," and "eutectoid" corresponds to "secondary.")

The Cored or Dendritic Structure

It will be seen later that solid solutions form in a certain way, calling for slow rates of cooling if the crystal grains which form are to be of the same composition (or concentration of components) throughout. If sufficient time is not given to this cooling, it often happens that the individual crystal grains are not homogeneous throughout, but are seen to consist of concentric shells of a strange shape often like a fern leaf, or an oak leaf; they seem to have an inside "core" surrounded by "shells" of different tints. This visible structure is so marked in many cases as to mask the usual grain boundaries and lead to the belief that the grain boundaries must follow the strange contours of the shells; sometimes they may do so.

CHAPTER VII

A SURVEY OF PREVIOUS CHAPTERS

As the practical objects of this Course are its strong features, it would be advisable to restate them shortly: they are to draw attention to the internal structure by which plain metals exhibit their mechanical properties, and to show how the mechanical properties may be developed in certain definite ways.

With those objects in view, attention has been drawn to the simplest metal of all—the liquid metal; to the next in simplicity—the metal crystallized as one single crystal (though this is the most uncommon metal of all); to the metal in the commonest known form—the polycrystalline mass, consisting of "amorphous" metal together with perfectly crystalline metal; and finally to the polycrystalline mass of metal which has been deformed at temperatures below the annealing temperatures, and which therefore contains "amorphous" metal inside the grains as a result of disruption and chaos at the gliding planes.

It has been shown in a preliminary manner how properties may be developed or modified as follows—

Methods of Modifying the Properties

(i) *Grain Refining* notably inducing high "shock" resistance, but both strength and ductility are retained or improved.

(ii) *Cold Work* induces elasticity, strength, and hardness, at the expense of ductility; fullest cold work means the entire absence of ductility in most cases.

(iii) *Alloying to form a Solid Solution* whereby increases in elasticity, strength, and hardness can be obtained which can be made to vary over a wide range by varying the amount and nature of the alloying metal; sometimes the resulting alloy has its ductility but slightly altered, sometimes the ductility is actually improved, though as a rule the ductility is reduced; sometimes corrosion resistance is improved, and in consequence fatigue resistance.

(iv) *Alloying to form a solid solution at the higher temperatures from which a compound may be precipitated at the lower temperatures,* whereby (with or without the effect of the solid solution formation as above) a compound is made to appear as molecules in the metal lattice. In this particular way it is possible to raise the strength, etc., to the highest possible degree for any given degree of ductility; the range of available alternatives by different heating and cooling treatments is greatest; the strength–ductility ratio is usually optimum (optimum toughness).

An Important Generalization

On application of force, metals change shape by sliding movements in each crystal of parts of the crystal over others; there is also flexure and flow of adjustment in the grain boundaries.

On the other hand, atoms are *not* lifted away from atoms in tension; atoms are *not* forced into atoms in compression (except to a slight extent quite elastically as when metal is subjected to uniform pressure all over).

In tension and in compression, as also in shear and in mixed loading, change of shape is therefore due to shear motion on the gliding planes of the crystal lattices, and to the accommodating flexure and flow of the boundary cement.

Therefore a metal's resistance to all kinds of loading must also be at the gliding planes, and in the resistance of the boundaries to flow.

This important generalization greatly simplifies the understanding of all the mechanical properties of metals. For the mechanical properties are the results of tests, and the tests are loadings; metals react to these tests at the crystal gliding planes, hence *the state of the gliding planes determines the mechanical properties* if the effect of the boundaries can be considered a constant and a minor factor.

The boundary cement has been considered in Chapter II; in the usual polycrystalline mass it accounts for some of a metal's strength at ordinary temperatures and at common loading speeds; but the quantity of boundary cement is not capable of being varied much in practice, for it is required to be in highest possible amount (finest grain-size) to obtain highest notch-toughness. Hence, as long as the presence and the nature of the boundary cement are not forgotten, their effect can be considered a constant factor while the effects of varying the gliding planes are studied.

The Next (an Intermediate) Stage of the Course. We have now concluded the first and reached an intermediate stage of the subject. Ultimately it will be necessary to illustrate the above methods of modifying the properties by reference to the actual metals and alloys in everyday use; but, in the meantime, it would be better to devote some attention to the mechanical properties themselves, to their characters and relationships. Without such a digression it would not be by any means so evident that the metallurgist's adjustments at the lattice gliding planes are responsible for all the modifications of properties intentionally induced in the metals; nor would it be so clear that accidental effects of heating and working can be explored at the gliding planes alone.

The following chapters, VIII, IX and X, are those of the intermediate stage of the subject; they deal with the mechanical properties themselves, and with other features concerning which engineers and metallurgists should see eye to eye. As long as an engineer can follow the metallurgist's technique of the "cooling curves" and the "equilibrium diagram," there is very little necessity to memorize it, and certainly no need to memorize it in any detail.

CHAPTER VIII

THE MECHANICAL PROPERTIES

IT is necessary to adopt a realistic attitude towards these mechanical properties, a forthright and restricted outlook upon each of them is essential if we are to get anywhere. The trouble has been that this technology relating to the mechanical properties has employed as terms so many words which bear vague meanings, words which mean different things to different people according to their own particular experiences and feelings. We should undoubtedly abandon all preconceived ideas, if necessary even our most favoured ones, to adopt in lieu the literal meanings and implications of the *actual tests* by which we obtain *figures of merit* for these properties, these latter being the "legal tender" throughout the world.

An illustration should make clear the need for being very careful and practical—it is useless to view "ductility" as the property by which a material be drawn out (like wire through a die) and to *think* of that property as involving some degree of "softness," when others will have somewhat different conceptions of the property; *more particularly as everyone* will actually measure ductility by the same test which certainly is not a test of drawing through a die, and does not involve any implications of softness or hardness. The test for ductility is the tensile test which supplies a "percentage elongation of the two-inch test piece at fracture." This, then, in its full simplicity, should be the engineer's and the practical man's chief definition and conception of ductility—the percentage elongation of the two-inch test piece before fracture.

"Strength," a cardinal word, is extremely vague until we decide what its measure is to be; as soon as we have decided what test to apply to measure it, we should accept the findings of that test, and all the implications of that test, to stand for these solidly in spite of any other notions on the subject.* By plain "strength" most people mean "ultimate tensile stress"—a serious misnomer which should really be called "*apparent* ultimate tensile stress"—but, since we know how it is measured, the misnomer does not matter so much. It is measured as the highest stress recorded in the usual way in the tensile test on the two-inch test piece; and the implications of this are extremely important if we are to understand these mechanical properties.

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* See "Stress Properties" at the end of this chapter.

Simple as this test for strength may sound, the facts of its nature have remained obscured; for the stresses of this test are habitually measured quite incorrectly; as are also the strains. So the resulting stress-strain diagram is at the best a convenient substitute for the real thing, and cannot even be described as an "approximation," because it is nothing like the real thing in tests on ductile metals.

And so we discover that a realistic, restricted, and forthright attitude means analysis first; then an acceptance of the real to the exclusion of the unreal; and, finally, very much simplified real conceptions which lend themselves naturally to explanations in terms of the resistance on the crystal gliding planes of metallic materials.

The Stress-Strain Diagram

Therefore, on the assumption that the understanding of the real meanings and implications of the standard tests is far more valuable than an insistence upon ideals or definitions for which test conditions cannot be conveniently reproduced, the most useful visualization of these mechanical properties will be obtained by scanning the "stress-strain" diagram.

This so-called "stress-strain" diagram curve of the tensile test actually supplies three of the figures of mechanical properties outright, and figures for others may be deduced as well; but the curve is not really a "stress-strain" curve: it is nothing but a load-extension curve to another scale, because in plotting it no allowance is made either for the change in the area of cross-section of the specimen which takes place as plastic distortion proceeds, or for the change of the effective length of the specimen as necking takes place. In this so-called "stress-strain" diagram, however, the curve corrected for stress and strain can be shown dotted where it differs from the load-extension shape.

A mere glance or two at such a curve (see Fig. 11) will demonstrate vividly—

(i) The *slope* of the first straight line part of curve is the *Elastic* Modulus. (This slope has been greatly exaggerated to show it.)

(ii) The maximum height of the curve is the *Strength*.

(iii) The maximum horizontal length of the curve is the *Ductility*.

The same Ductility as percentage elongation may be due to—

(*a*) Less plastic deformation over a longer length of the test piece, or (*b*) more plastic deformation over a shorter length of the test piece. In the former case the "*Reduction of Area*" at this point of fracture is less, and in the latter it is more.

Reduction of Area, expressed as a percentage reduction of the cross-sectional area of the test piece, is a common and useful specification by which to *qualify* the basic (percentage elongation) test for Ductility. The lower the R.A. the more the metal will strain-harden under cold work; the higher the R.A. the less will it tend to strain-harden, and the easier to fabricate in the cold. In the common stress-strain diagram, the *drooping* of the curve beyond the point of measurement of the Strength (U.T.S.) indicates the extent of the R.A.

(iv) The area under the curve (being the energy required to fracture) is a measure of the *Toughness*.

The greater speed of the loading in the Izod test usually employed to measure Toughness makes no appreciable difference to the energy required to fracture, whereas the notch may make a considerable difference; moderately fine notches in this test discriminate between metals according to what is called a "*Notch Toughness*."

(v) In the case of the ductile metals, *True Breaking Stress* may be nearly three times the Strength; but as lack of ductility is manifested, it becomes correspondingly smaller multiples of Strength, ultimately being the same as Strength for the inductile metals.

(vi) *Limit of Proportionality*, *Elastic Limit* and arbitrary *Proof Stresses* (such as the 0·1 per cent, the 0·2 per cent and the 0·5 per cent) can be seen in the diagram. The latter are sometimes described as *Yield Strengths*. Some metals in an exceptional state exhibit fixed Yield Points, but this behaviour is not characteristic of the stress-strain curves of the metals in general.

With regard to the above, it should be explained that in the case of many metals the "primitive" or ordinary elastic limit is ill-defined, and is found to be lower and still lower as methods of measuring strain become more and more refined. It is therefore found desirable in such cases to apply a test to find the stress producing no more than a fixed amount of permanent set (such as 0·1 per cent, 0·2 per cent or 0·5 per cent) after the load inducing it has been on for

15 seconds only. This is the Proof Stress or Yield Strength of an arbitrary standard. In practice a specimen passes the test for Proof Stress if, on inducing this proof stress for 15 seconds, the permanent set is not in excess of the fixed percentage chosen.

(vii) The following mechanical properties cannot be so well visualized from the stress-strain diagram—

Hardness by the indentation tests (such as the Brinell) will be found in essence to reflect the Yield Strength under working conditions of a particular nature. Upon analysis, the mechanics of the indentation tests prove to be such that the stress measured as the Hardness "number" is a nearly constant multiple of the effective Yield Strength. The well-known ratio—

$$\frac{\text{U.T.S.}}{\text{B.H.N.}} = 0{\cdot}22 \text{ (approx.)}$$

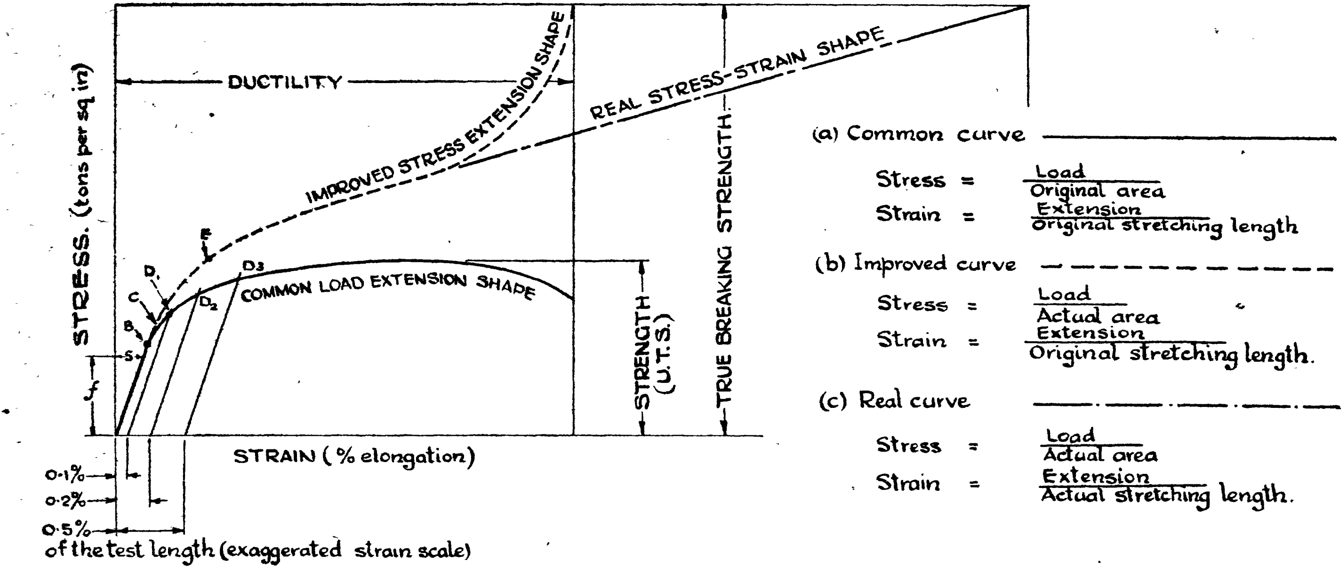
is of undoubted convenience for approximating to the Ultimate Tensile Stress (Strength) in many cases; but it is not a real or significant ratio by any means, being incidental only to the real significant relationship above mentioned.

"Creep" in Metals

None of the above strength measurements at the tensile test rate of loading is suitable for testing metals which are required to retain dimensions under stresses for very long periods, as for example in the case of pipes, bearings, steam and chemical plant, furnaces, or turbine blades, etc.

At raised temperatures the elastic limits and proof stresses determined at ordinary temperatures do not hold good; but this is not all, for *strength and ductility appear to be functions of the loading speed or stress induction rate*, so that a metal is stronger and less ductile under fast loadings, and weaker but *more ductile* under slow loadings. The grain boundary "amorphous" cement resembles pitch which flows more under light loadings, though it stands up to much greater loadings and fractures under them. It is found that fine grain in a metal—more boundaries—means more creep.

Many metals and alloys are now carefully tested for creep at the temperatures they will attain in use; strain in millionths of an inch per inch length per period (say 100 hours) is determined at different stresses and at different temperatures in simple apparatus using sensitive dial or optical extensometers.



(a) Common curve ————

Stress = $\frac{\text{Load}}{\text{Original area}}$

Strain = $\frac{\text{Extension}}{\text{Original stretching length}}$

(b) Improved curve – – – –

Stress = $\frac{\text{Load}}{\text{Actual area}}$

Strain = $\frac{\text{Extension}}{\text{Original stretching length.}}$

(c) Real curve —·—·—

Stress = $\frac{\text{Load}}{\text{Actual area}}$

Strain = $\frac{\text{Extension}}{\text{Actual stretching length.}}$

A.- Elastic Modulus $\frac{f}{s}$ (grossly exaggerated slope)
B.- Limit of Proportionality
C.- Elastic limit
D_1 D_2 D_5 - Proof-Stresses – Arbitrary Yield Strengths.*
E - Average true Yield Strength of value approx. that of the U.T.S.

* Fixed yield points are sometimes to be found but these are not characteristic of the Stress-Strain curves, and are probably incidental to some peculiar condition of the metal.

STRESS - STRAIN DIAGRAM.

FIG. 11

Fatigue Resistance

Metals in which simple, fluctuating, or alternating stresses are repeatedly induced are found to fail at stresses lower than the ordinary ultimate stress of the relevant static test. Most of the repetition stressing induced in practice may for convenience be analysed into some *mean static stress* upon which is superimposed an alternating stress of a certain *range* (plus and minus). At the zero extreme of mean stress, the ultimate alternating stress-range has a maximum value. When stress-range is the ultimate (causing or near fracture) in as many as ten million alternations or reversals, it is called the "safe range"; and the safe range at zero mean stress is called the "fatigue endurance range" or "fatigue limit," a measure of a recognized property of materials called the "Fatigue Resistance."

The upper extreme of the static mean stress is, of course, the relevant static ultimate stress itself—at which the safe stress-range must be zero.

The Goodman Diagram was drawn up to show the safe ranges of alternating stress to be expected for different values of the mean stress, or vice versa, if the relationship between these two is of the simplest or linear kind.

Gerber has suggested a certain more complex relationship between safe range and mean stress; but his formula and ideal curves are not followed by experimental results very much more closely than those which may be derived from the Goodman Diagram. Experimental results in general cannot be said to confirm either of these suggested relationships in an emphatic manner.

Fatigue Testing

The favourite tests to supply figures of merit for the fatigue-resisting capacities of the metals are those subjecting a test piece to simple alternations of tension and compression, or to equal reversals in torsion, to ascertain the safe range at zero mean stress. Such tests are not carried out upon one test piece starting at zero stress-range and working up to the stress-range at fracture point, for fatigue in actual practice never follows such a course; the result of such a test is always found to be unduly optimistic, for reasons touched on later.

The full test should be carried out with numerous specimens of the same material, each with a different stress, and to each of which is applied its own stress-range continuously from beginning to end.

The end, however, may either be fracture at some specific number of reversals less than ten million, or fracture at about ten million, or no fracture even at ten million reversals.

From a curve drawn to exhibit the relationship between stress-range and the number of reversals to cause fracture, a fatigue limit stress-range usually becomes evident from a flattening out of the curve at or near the ten million reversal mark. This is of a magnitude which, in about half the specimens tried, will not quite cause fracture; and which, in the case of about half of the specimens, will only just cause fracture, in ten million reversals.

So much for the more essential facts of fatigue in practice and in testing. What do these really mean? We can understand why safe stress-range must depend upon static mean stress; but it is more difficult to see why, at zero mean stress, the alternating safe range is so much lower than ultimate static stress under specimen-loading of every type.

The Analysis of Fatigue

The fatigue limiting stress is low in comparison with some other stress which we take to be a fairer standard of strength; but the difference can only be due to the difference in the manner of testing, and therefore will be explained as soon as the analyses of the static tests and of fatigue tests have been carried to the necessary point of elucidation. Assuming that the analysis of the static tests has gone far enough, it remains to answer the riddle by a close analysis of some straightforward type of fatigue such as that in the Wohler test. The answer must be found somewhere in correct interpretations of the effects peculiar to fatigue—rapid loading, repeated loading, crack propagation, special fatigue-type stress-raisers, and so forth; but this answer is not yet forthcoming.

Satisfactory in many respects are the assumptions that the stress properties at the fatigue rate of loading are higher than those at the ordinary tensile test rate, that the fatigue limits are within these high elastic limits, but that they may be very much lower than these on account of the disastrous effect, under repeated reversals, of a special type of stress-raiser. On these assumptions, the explanation of fatigue failure becomes little more than the elucidation of that particular type of weakness or "inhomogeneity" in the metal which is the fatigue stress-raiser. It would only be necessary to discover—for the sake of example—that certain localities in a metal must become strained more in the compressive quarter-cycle than in the tensile quarter-cycle (above a certain critical stress) to show how local straining of this kind would produce cracks and ultimate failure. Fatigue would then be a mystery no longer.

In general, it seems to be the function of the fatigue test to exhibit and to measure the fatigue-type stress-raiser; but, before this can be done, it is necessary to know what the ideal fatigue limit is. The metal's ideal limit would seem to be its true fatigue-rate breaking stress, though little interest has been taken in forecasting or hazarding guesses as to what stress would be reached in the entire absence of these stress-raisers.

Goodman attempted to show that, in such cases of load reversals at speed, a dynamic effect may be such as to cause peak stresses of as much as three times their measure as static stresses; in other words, when we think we have induced alternating stress of (say) 20 tons per sq. in., in reality we may have induced one as high as 60 tons per sq. in. Little credence is now given to the existence of any dynamic effect in testing by the Wohler type of machine, but some interesting features of the fatigue phenomenon can be brought to light in a serious attempt to apply Goodman's suggestion.

The known fatigue limits multiplied by three (Goodman's dynamic factor) would in most cases be far beyond the elastic limit, but any idea that the limit stress must be less than the common or primitive elastic limit is erroneous. The fact that the fatigue specimen retains its shape and dimensions up to the final crack does not in itself argue any lack of plastic strain, but perhaps only a compensation of plastic strain—an equal amount of slip in both directions whereby the original size and shape are retained. It is true that there may be very little plastic strain exhibited in the case of some specimens fractured in fatigue.

In a very large number of cases, however, three times the fatigue limit is a stress in excess, even, of the U.T.S. Yet, here again, the (nominal) U.T.S. really has nothing to do with fatigue in spite of the curious roughly constant ratio between the limit and the U.T.S. which has been misleading and is still not understood. The U.T.S. cannot possibly be the strength criterion of a specimen which has retained its original shape and cross-sectional area; a real ultimate stress of fracture might be the figure to employ as the strength criterion, and the real ultimate stress may be several times the nominal.

When referring to these tensile test properties it is as well to remember, also, that values of stress properties will increase with the rate of loading; that the speed of loading in the fatigue test is very high; and that the real ultimate fully strain-strengthened stress at the fatigue test's high rate would be even higher than the real tensile ultimate stress.

If these fatigue tests are carried out in the wrong way by stressing gradually from zero stress-range up to fracture stress-range, we always find higher fatigue limits—possibly coinciding, when multiplied by three, with the real ultimate (fatigue-rate) fracture stress according to the requirements of a dynamic loading effect. But, in the case of tests which are properly carried out to reproduce practical conditions, the stress would not rise so high; for the stress of the correct and safe range (at zero mean stress) is slammed on the specimen from the beginning, with the result that the strain-strengthening proceeds at a very rapid rate; at so rapid a rate, in fact, that failure in the form of tiny cracks will take place at stress-raiser points where stresses are unduly raised locally unless radial strain distribution may spread out from these points at the same rapid rate. Such points of weakness are present in large numbers; even in very carefully prepared specimens of the best metal it is to be presumed that there are minute superficial and internal blemishes, scratches, corrosion pits, and junctions of phases, all of which act more or less powerfully as stress-raisers in their own localities.

According to one dynamic theory of fatigue, then, the fatigue limit would be one-third of the true (fatigue-rate) breaking stress less some amount which depends on the damaging effect of the fatigue type stress-raiser. In general, if a dynamic effect of any kind or magnification factor is present in fatigue testing, the fatigue limit would prove to be some fraction (depending upon the factor) of the true fatigue-rate breaking stress less the effect of stress-raisers in terms of stress.

"Corrosion Fatigue"

If proof were required of the dangerous nature of tiniest invisible flaws and scratches, it is only necessary to compare the fatigue limits of specimens tested under corrosion conditions and under corrosion-free conditions. Invisible corrosion pits at the surface of the specimen are found to influence the fatigue limit enormously, showing how readily the test detects the mechanical and metallurgical weaknesses responsible for failure under all kinds of repeated stresses.

Fatigue failures in testing and in practice under many different conditions of use always exhibit on the fractured surfaces signs of trouble in the form of a crack spreading gradually from some local centre of weakness until the remaining cross-section of the specimen or of the part in use breaks down suddenly under gross overload.

Damping Capacity

Descending to further details of repetition loading, it is found that metals and alloys differ in their response to stresses beyond the ideal elastic limit, beyond the "limit of proportionality" within which induction and removal of stress follow a perfectly straight line from the origin of the stress-strain curve.

As soon as stresses have exceeded the above limit there is a slight lag of strain removal behind stress relief, and subsequently a corresponding lag of strain behind stress induction, so that a hysteresis loop is formed on repetition of the stressing. This means an absorption of energy corresponding to the area of the loop. As increased stressing followed by unstressing is carried out, these loops become wider and longer, and increasingly great amounts of energy are thereby absorbed.

Metals vary in the area of these hysteresis loops, and therefore offer *differing capacities to the damping out* of any loading which tends to produce energy oscillations in them; here is a property determining whether metals will or will not permit the building up in them of vibrations which might be dangerous under undamped resonating conditions.

Iron and mild steels are peculiar in the respect that long periods of rest, or short periods of very mild heating between the stress inductions, will restore new limits of proportionality to the steels; cast iron happens to have a high damping capacity to make up for some of its deficiencies in other respects.

Stability and Long Life

Now and again the complaint is heard that, whilst engineers really want information as to how a metal will stand up to its job, all metallurgists can offer instead is a string of mechanical properties which do not supply the required information, and cannot be interpreted to supply it.

The mechanical properties have to stand for a very large and varied number of possible uses: they are probably the minimum and as simple as possible in consideration of all these uses.

The elastic modulus is required wherever resilience calculations must be made.

If simple direct loading in tension or compression is to be provided for, metals are expected to yield not at all or merely negligible amounts; hence no stresses beyond the elastic limit or proof stress can be considered, so that these limits must be registered with accuracy by means of sensitive devices in the tensile test.

As engineers are somewhat rarely concerned with the nominal breaking stress (long before the attainment of which the metal may have changed its dimensions considerably), they are liable to discredit the value of the U.T.S. Quite apart from the rare occasions on which the U.T.S. may be directly useful with some small factor of safety, it is a simple test the figure of which commonly bears an ascertainable relationship to the fatigue limit which no one can possibly ignore, but which is much harder to test. For steels of different strengths the fatigue limit may be approximated by applying factors to the U.T.S. varying from about 0·40 (for the highest tensile steel) to about 0·55 (for the lowest). Moreover, wherever static stresses must occur where fluctuating and alternating stresses will also occur, it is customary to design from experience on a figure of U.T.S. with a factor of safety of about 4 for the static stress, of about 5 for the fluctuating stress, and of about 6 for the alternating stress.

The wearing properties of a surface is, perhaps, sufficiently measured by an indentation test like the Brinell; such a test is the quickest of all to make, and its figure bears constant relations to the U.T.S. and to the fatigue limit for any particular metal. An indentation test is, then, the quickest and most valuable one if fully interpreted.

Where structures are liable to pure alternations of stress, the fatigue limit is probably the most valuable single test figure of all. Most cases of repeated stresses in structures are not, however, pure alternating stresses like the fatigue test itself, so that mean stress, as well as stress-range, is vital. It is evident that stress-range must decrease with rise in mean stress, until it is zero when mean stress reaches the U.T.S. Here again, the U.T.S. proves to be necessary as a *basis of calculation in design*, though it would not do to employ it in this or in any other calculation without a sufficient factor of safety.

The Factor of Safety

This is more correctly viewed as a "factor of uncertainty." Factors of 4, 5, and 6 may seem large ones to use after all the attention paid to the preparation of metals, and to their tests; but it should be remembered that the basis of calculation, the U.T.S., is about twice the elastic limit as a rule, and metals and their tests are by no means the only uncertainties either. Metals and alloys are getting more reliable according to evidence of very much improved testing, but it cannot be said that the problems of the *designer* are becoming simpler in the internal combustion engine or in the aeroplane, nor can it be said that the uncertainties *in the use* of these machines are no more considerable than in the cases of the steam engine or the bridge.

Metals, tests, design, and use (the latter including the bugbear of live-loading to double and treble the stress) all call for margins of safety: it is not very surprising that the overall margin, or Factor of Safety, is large.

Ductility

Of all the mechanical properties this has seemed to the engineer to be the most unnecessary. Seeing that new structures ought to be able to settle down with local extensions not exceeding 0·1 per cent, and that any structure must be rendered useless by a distortion of 1 per cent, it seems futile to argue the relative merits of metals with high ductilities which can never be made use of.

Apart from the occasions on which ductility of a high order can serve as a measure of a metal's capacity for cold or hot work, ductility *as high as is compatible with* more essential properties is called for as a safeguard, as a purely metallurgical factor of uncertainty, to look after some doubtful features of metals and of their tests. High ductility is commonly a sure proof of a metal's uniformity and cleanliness at the grain boundaries: a steel, for example, which is not exhibiting its usual high degree of ductility is evidently suffering from a metallurgical defect which may *indirectly* affect its behaviour in use for some purposes, including structural purposes not calling for high ductility itself directly. In short, if a metal can be made to exhibit a certain high ductility, and yet does not do so, there is something the matter with that metal which might prove disastrous in the unforeseen end as a result of corrosivity, creep, notch sensitiveness, poor damping capacity, or general lack of consistency between one specimen of the metal and another.

Until comparatively recently the highest class crankshaft material was steel made in the usual way which in fabrication is subjected to elaborate forging; and it was possible to have ductility as high as 60 per cent in such steels as were considered otherwise fit for crankshafts. In some chosen steel which, on proper heat treatment, displayed such high ductility, ductility would undoubtedly be an important criterion of the general efficacy of the metallurgical processes involved in its preparation.

Latterly, a crankshaft has been made in quite a new way from an alloy cast iron by very special heat treatment; the steel-making stage has been cut right out, and no forging is called for. Now, the inductility of this product does not indicate anything deficient in the manufacture or preparation for use, because it proves to be quite impossible to give this metallic material more than a mere trace of ductility. But this new crankshaft has all the desired essential properties, and proves to be as consistent in its behaviour as any good uniform metal can be.

Stress Properties

"Strength" is a word often employed with a far more general application to all properties measured in terms of stress; whereby hardness, fatigue limit, modulus of elasticity are all included as particular kinds of strength. It would be of convenience to have some such expression as "the stress properties" by which to refer to all these different "strengths" or properties measured as stresses; for, in general, all such properties change in the same sense, rising and falling in unison whenever anything is done to the metallic structure causing any one of them to change. Hardening, for example, is a process which achieves much more than an extra resistance to abrasion and penetration: hardness is a stress property, and where and when you produce greater hardness you are bound also to produce rises in all the other stress properties.

Strain Properties

Similarly, it would be convenient to group all those properties which are actually measured in terms of strain—or which would be so measured if they were measured at all—as "strain properties." These would include ductility, creep, malleability and extrudability; they vary in unison inversely as "stress properties."

CHAPTER IX

THE METALLURGIST'S TECHNIQUE

The X-ray Examination of Metals

ETHER waves or electromagnetic radiations, like light and radiant heat but of much shorter wavelength, can be made to emanate from the "anti-cathode" of the "Cathode Tube"; so much shorter in wavelength is this "X" radiation that it penetrates objects opaque to light, even metals for some depth, in a manner suitable for two kinds of investigations—

Radiography

The method whereby bone dislocations, fractures, or metal objects lodged in the human body may be recognized and examined by X-ray shadow-throwing is familiar to all; cursory examination of all such shadowgraphs may be made by the use of a fluorescent screen—a plate of glass coated with a substance like barium platinocyanide which is excited to give off light when X-rays (themselves invisible) impinge upon it.

But X-rays also photograph themselves on ordinary films for permanent record just like light rays.

X-ray shadow examination by the fluorescent screen and photographic film is known as Radiography: it is regularly and usefully employed nowadays to a tremendous extent for routine and research investigation into the soundness of metal castings, forgings, and stampings.

X-ray Analysis and Spectra

A quite different use of X-rays is that which yields information about the internal structure of metals, about grain-size and internal stresses.

Unless special precautions are taken, X-rays of mixed wavelengths emanate, just as light from any casual illuminating source may be of many different wavelengths or colours. It is possible, however, to have monochromatic light (such as the yellow sodium light) and it is also possible to have the equivalent thing in X-rays—a "monochromatic" X-ray beam.

If a fine perfectly straight "monochromatic" X-ray beam is made to strike a metal, it penetrates some distance into the metallic crystalline aggregate but much of it is *reflected backwards* out through the metal again in straight line paths *at certain definite angles* with the incident beam. Upon films situated (behind the incident beam) to catch such reflections, *ring-shaped images* may be photographed, the measurements of which show the distances between the units of the metal space lattice, and a great deal more besides.

It is beyond the scope of this account to give more precise details of the investigations; but it may be said that familiarity with the above technique has proved its extraordinary accuracy in determining the distances separating the units on the different planes of the lattice, in observing the progressive effects of cold work, and (recently) in estimating the situation, amount and direction of internal stresses in a metal.

Chemical and Spectrographic Analysis

Chemical analysis for the determination of the exact composition of metals and alloys has been developed to a high degree of accuracy, and is still the chief method employed. In recent years, however, developments in the study of the spark spectra of metals have made possible an accurate, rapid and inexpensive method of quantitative analysis—without the use of chemical reagents—by spectrographic means alone; the popularity of this new analysis is growing rapidly.

The Microstructure of Metals

By means of the microscope with magnifications varying from 30 to 3000 the microscopic and coarse structures of metals and alloys can be inspected. Sub-microscopic structures are often deduced from these, particularly when X-rays have shown what kind of deductions may be drawn.

Preparation of Specimen

The specimen to be observed has one face very highly polished, commencing with comparatively coarse emery paper and working down to a very fine emery paper. The final polish will be given with cloth and the finest abrasive.

Etching

The specimen is then etched with a suitable reagent which is chosen, in the case of an alloy, so that it will react with the constituents present to distinguish them. In this way certain portions will be more acted upon than others, and consequently we shall have light and shadow, bright and dark areas, thus enabling the distribution of the constituents to be seen.

In the case of a pure metal, owing to different orientation the crystal grains will etch to different extents and will also differ in their capacity to reflect light upward to the eyepiece; shadows or black lines will show up the grain formation. If the specimens are to be used again, they must be left in a dry atmosphere or protected from corrosion by some means such as a wax film, or wrapping in "Cellophane."

Heat Treatment and Equipment

The alloy, if susceptible, is subjected to a process of heat-treatment by means of which it is possible so to arrange the constituents of the alloy, and their internal structure, that the most suitable mechanical properties are obtained. To heat-treat the larger number of different alloys in use it is necessary to have apparatus giving a range of temperature from 250° C. to 1300°.

This range is covered by the use of—

Salt baths which are gas or oil heated—250°–1000°.

Furnaces usually gas heated—up to 1300° C.

Electric furnaces, A.C. and D.C., heated by elements up to 1300° C.

Salt baths electrically heated by immersion plates up to 1300° C.

Measuring Temperatures up to 1200° C.

For this purpose Pyrometers are used. The pyrometer is usually a thermocouple consisting of two wires of different metals or alloys. The two wires are twisted or welded together at each end. If one junction is subjected to heat, while the other is kept at some constant low temperature such as that of melting ice, a potential difference is established between the junctions. This varies as the difference in temperature between the junctions. A milli-voltmeter, the scale of which is calibrated in degrees of temperature, is inserted into the circuit.

When properly protected by a suitable porcelain sheath, a couple having one wire of pure platinum and the other of 90 per cent platinum with 10 per cent rhodium can be used for temperatures up to 1400° C. Such a couple develops an E.M.F. of about 0·01 volt (10 millivolts) at a temperature of about 1000° C.

For continuous work the couple should not be used for temperatures above 1200° C.

Measuring Temperatures up to 800° C. or so

One wire is nickel and the other an alloy of nickel and chromium. The E.M.F. produced is approximately four times as high as that in the case of the platinum and platinum-rhodium couple.

A sheath of steel or nickel chrome alloy is used to protect the couple from oxidization and corrosion.

Measuring Temperatures over 1200° C.

Where readings are only necessary from time to time an optical pyrometer is used. This apparatus consists of a tubular eyepiece in which is fixed a small filament lamp. The lamp is connected to an accumulator via a small adjustable rheostat and a milliammeter calibrated to read in degrees of temperature. The observer looks through the eyepiece at the substance, the temperature of which is to be measured. The rheostat is adjusted until the filament just disappears from view, and the temperature is read on the milliammeter. The filament will disappear when its temperature is exactly that of the substance being viewed.

For continuous reading of temperatures over 1200° C. it may be convenient to use a radiation pyrometer which works by focusing the heat rays given off by the hot substance on a thermocouple situated at a distance from the substance.

The Equilibrium Diagram

Neither the scientist on the one hand, nor the engineer on the other, can appreciate the technology of the metallurgist without first grasping the elements of the system which has proved so fertile in the investigation of the alloys, and resulted in unparalleled advances in the output of engineering materials.

The physical metallurgist plots experimental results with a regular sequence of alloys between 100 per cent A metal and 100 per cent B metal; he finds, in the case of each composition of the sequence, the structures at a sequence of temperatures from high to low (from the completely molten state to the solid state at room temperatures).

After such an investigation lines, or rather a series of lines or curves, are obtained in a diagram called "The Equilibrium Diagram" (or temperature-concentration diagram, or constitution diagram), upon the completion of which the metallurgist can predict by interpolation with regard to any slowly cooled alloy of A and B, at an temperature—

(i) whether it is liquid, solid, or not quite solid;

(ii) the composition and proportion of the solid part; also those of the still liquid part, if any;

(iii) the solid constituents, their number and character (pure metals, solid solutions, compounds, eutectics, eutectoids, and the proportions of these);

(iv) melting range of temperatures;

and a good deal besides.

Strictly, these diagrams refer only to alloys cooled down slowly under ideal conditions for the establishment of equilibrium within them; but nevertheless they can be interpreted with ease to show the condition of alloys cooled at faster rates, or actually quenched from some particular temperature.

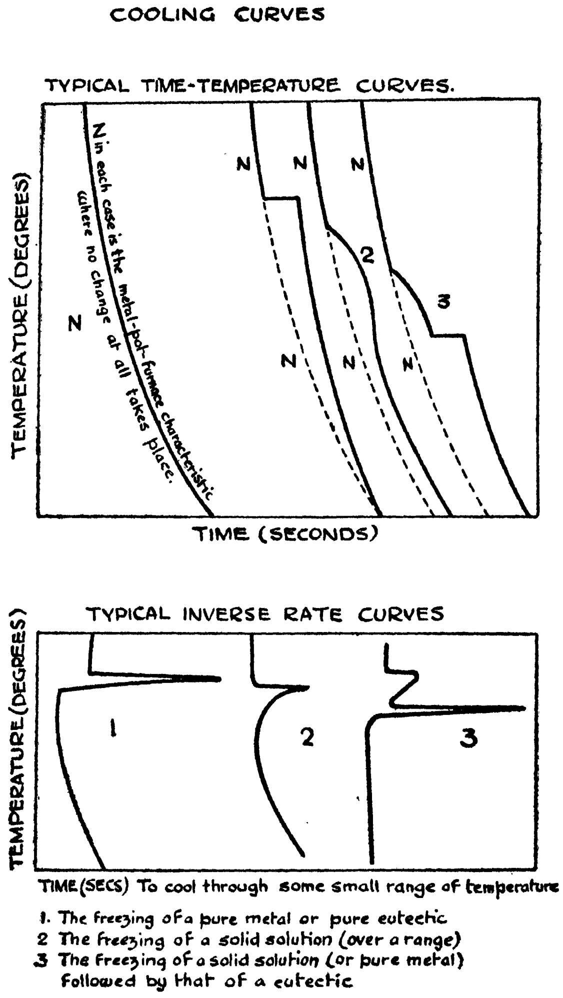
Cooling Curves

To investigate a series of alloys between A metal and B metal, cooling curves are taken of a chosen series such as 10 per cent B in A, 20 per cent B in A, 30 per cent B in A, etc., up to 90 per cent B in A (which is the same as 10 per cent A in B).

Cooling curves may be simple charts of *Temperature* (degrees) and *Time* (seconds), which are not very sensitive to slight heat evolutions or absorptions; or else they may be of a more sensitive type, an example of which is the so-called "*Inverse Rate*" curve. In the latter, the Time (seconds) required to heat or cool through a certain number of degrees of Temperature (say three degrees) is plotted against the actual temperature itself, when much slighter deviations from the perfectly normal heating or cooling rate may be exhibited than in the case of the plain Time-Temperature plotting.

When no changes of state or of solubility whatsoever occur during the heating or cooling of a metal, the curves have certain simple characteristics depending upon the rate of gain or loss of heat throughout the temperature range. Superimposed upon this normal shape, there will be changes of shape consequent upon any fusions, freezings, changes in liquid or solid solubility, or changes in allotropic form; and it is, therefore, the *changes effected* in the normal shape of the heating or cooling curves which characterize the interesting effects, and which with experience can be interpreted with ease to indicate what is happening.

In the diagrams will be seen examples of the two kinds of cooling curves, the normal characteristic of which is a sweep *down* from left to right; the heating curves (not shown) would be similar except for a sweep *up* from left to right.


COOLING CURVES

TYPICAL TIME-TEMPERATURE CURVES.

TYPICAL INVERSE RATE CURVES

TIME (SECS) To cool through some small range of temperature

1. The freezing of a pure metal or pure eutectic
2. The freezing of a solid solution (over a range)
3. The freezing of a solid solution (or pure metal) followed by that of a eutectic

FIG. 12

Each alloy is cooled down very slowly from a temperature well above its melting point with a thermocouple hot-junction in it—

preferably stirring it. Temperature "arrests" or "halts" will be indicated at various changes of structure from liquid to solid, at the conclusion of solidification, at the freezing of the eutectic, and also at any change of solid structure which may take place in the solid state indicating the breaking up of a solid solution, and the formation of a eutectoid.

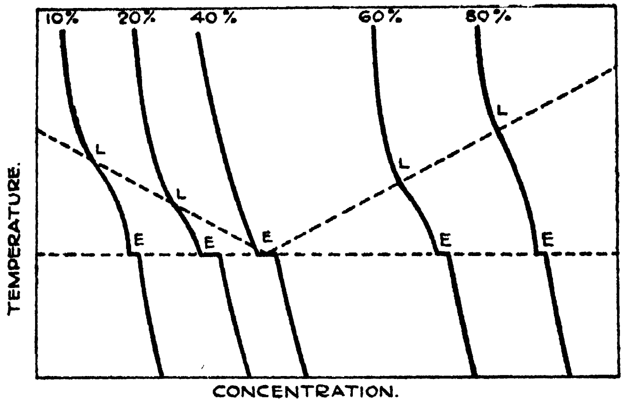
The arrest points in cooling for the whole series of alloys are plotted between a vertical ordinate of temperature and a horizontal ordinate of composition (100 per cent A to 100 per cent B); points representing the same type of structural change are then joined, and the story of the behaviour of the series which comes to light can be utilized to an extent but poorly indicated in the above superficial remarks.

CHAPTER X

THE EQUILIBRIUM DIAGRAM—GENERAL

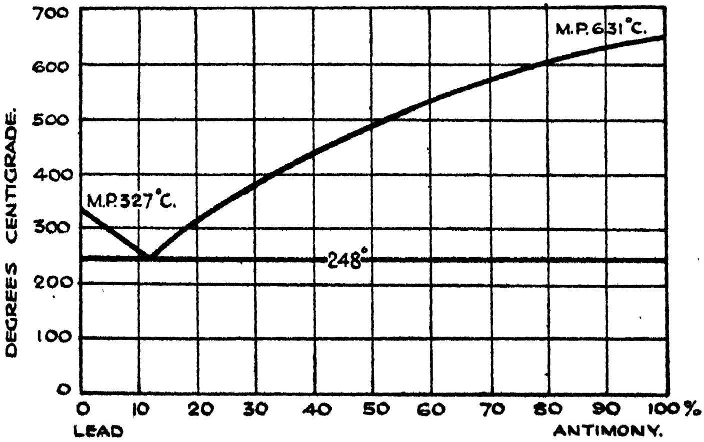
IN the diagrams can be seen the effects upon the time-temperature and inverse rate cooling curves of various structural changes; a

EQUILIBRIUM DIAGRAM IN THE MAKING.



SHOWING HOW THE EQUILIBRIUM DIAGRAM IS CONSTRUCTED FROM A NUMBER OF COOLING CURVES.

THE FINISHED DIAGRAM.



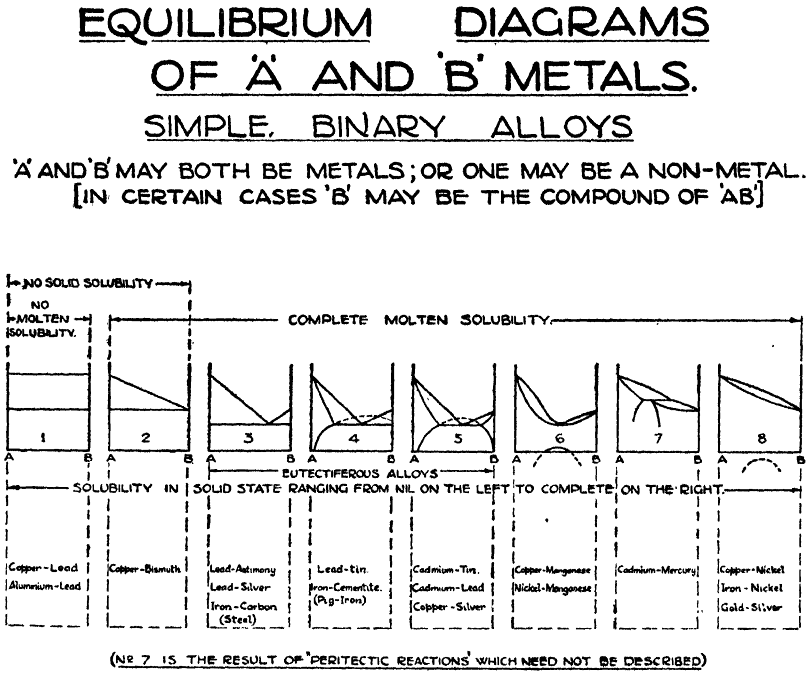
THE LEAD ANTIMONY DIAGRAM.

FIG. 13

fair idea can also be gathered of the way in which the cooling curve arrests are put into a diagram, giving rise to the lines of that diagram.

Notable among such lines are the "liquidus," a line above which the alloy is entirely liquid; and the "solidus," below which it is entirely solid.

The simplest alloy diagram is that in which no solid solubility


EQUILIBRIUM DIAGRAMS OF 'A' AND 'B' METALS.
SIMPLE, BINARY ALLOYS
'A' AND 'B' MAY BOTH BE METALS; OR ONE MAY BE A NON-METAL.
[IN CERTAIN CASES 'B' MAY BE THE COMPOUND OF 'AB']

(Nº 7 IS THE RESULT OF "PERITECTIC REACTIONS" WHICH NEED NOT BE DESCRIBED)

FIG. 14

is shown between the A metal and the B; such alloys (the lead-antimony alloy is an example) are eutectiferous, that is, consist of a primary metal (A or B as the case may be) with an AB eutectic—that is, unless the alloy is of the eutectic composition, when primaries will be absent and the whole alloy will consist of the one "constituent," the eutectic alone, the fine mixture or emulsion of secondary A and secondary B.

The intervention of a chemical compound AB at some specific composition would lead to a splitting of the one diagram into two, the A to AB compound diagram, and the AB compound to B diagram.

The next diagram in order of simplicity is that in which there is a limited solid solubility: metal A dissolves a limited quantity of B at any particular temperature, and B dissolves a limited quantity of metal A; but there is not an unbroken sequence of solid solutions between A and B.

Such alloys will vary as follows: up to the percentage of the maximum solid solution of B in A, the alloys will consist of one phase only, a homogeneous solid solution to all appearance exactly like a pure metal if equilibrium conditions have prevailed during the cooling. If not, the crystal grains will be "cored" or "dendritic" in structure, showing lack of uniformity in structure from centres to outsides, and an unduly low average percentage of B, so that in extreme cases an excess of B may be free to produce further constituents like eutectics or compounds. (See Chapters VI and XX.) Up to the percentage of the maximum solid solubility of A in B (at the other side of the diagram) exactly the same state of affairs will exist (except that A in B solution will not resemble the B in A solution in colour or other characteristics).

In between these two extreme compositions the alloys of A and B will be eutectiferous with a eutectic which will consist of a mixture of the maximum solid solutions of B in A and of A in B. Here again, there will be one alloy of eutectic composition which will consist of eutectic only.

Changes in the entirely Solid State

Whenever it appears, the eutectic is a "last freezing" constituent forming on the whole or on part of the solidus line; but the solidus does not necessarily finish off the diagram: the eutectic is by no means always the result of the last change.

It may come as a surprise to find that there is a mobility in the atoms of a solid alloy permitting certain changes of structure to take place—in the completely solid state—without conspicuous hindrance at the higher temperatures, though with increased sluggishness at the lower temperatures.

In many cases the solid solubility limits alter (usually falling, but often rising) with fall in temperature; and, in consequence, other lines of change will appear in the diagram extending below the solidus right down to room temperatures (and below if they are worth following lower).

Each of these lines is plotted as before by the indications of arrests in the cooling curves.

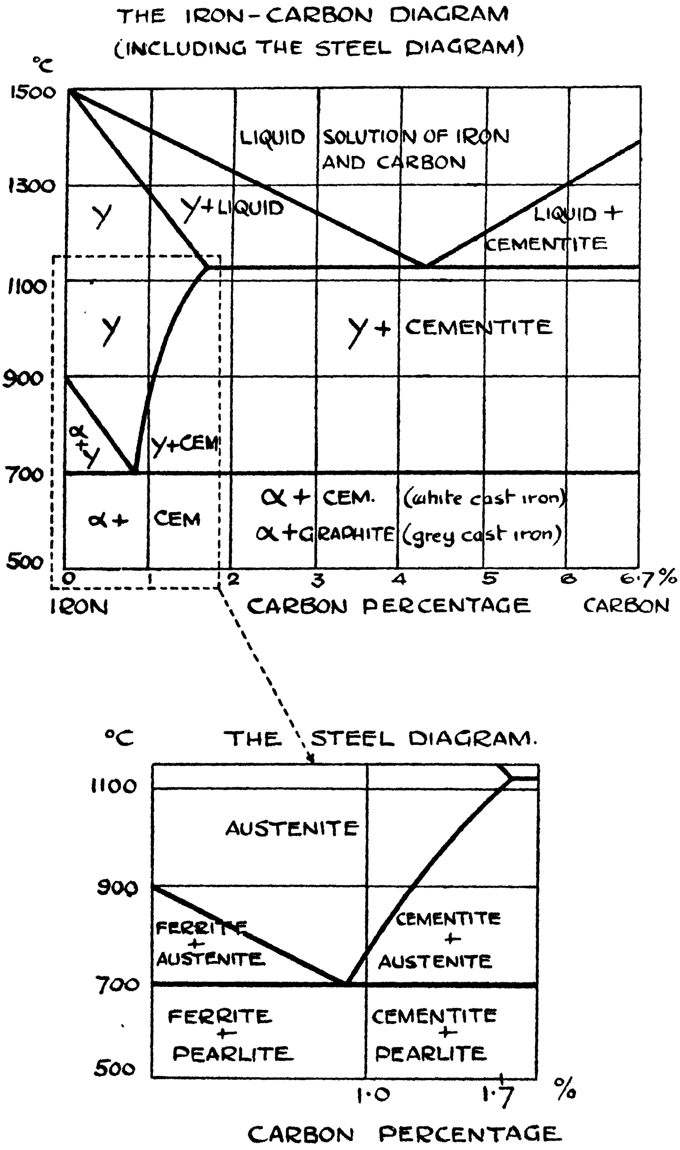

THE IRON-CARBON DIAGRAM
(INCLUDING THE STEEL DIAGRAM)
°C
1500
1300
1100
900
700
500
LIQUID SOLUTION OF IRON AND CARBON
γ
γ+LIQUID
LIQUID + CEMENTITE
γ
γ + CEMENTITE
α + γ
γ+CEM
α + CEM
α + CEM. (white cast iron)
α + GRAPHITE (grey cast iron)
0
1
2
3
4
5
6
6·7%
IRON
CARBON PERCENTAGE
CARBON
°C
THE STEEL DIAGRAM.
1100
900
700
500
AUSTENITE
FERRITE + AUSTENITE
CEMENTITE + AUSTENITE
FERRITE + PEARLITE
CEMENTITE + PEARLITE
1·0
1·7
%
CARBON PERCENTAGE

Fig. 15

The Iron-Carbon Diagram

The Iron-Carbon Diagram is shown: it deals with pig iron and cast-iron, in fact with every alloy of iron and carbon of any interest, including the carbon steels.

This diagram relates to the alloy of iron with a non-metal, carbon, but the result is just the same. It is an example of a series between 100 per cent iron and 100 per cent carbon which had been *interrupted* by the appearance of a compound of iron and carbon, Fe_3C, at 6·7 per cent carbon. Beyond this compound the diagram is of no interest: the engineer would sometimes like to discover what he can about cast-iron, but for the most part his interest will be confined to the region of carbon steels between 0 per cent and 1·7 per cent of carbon.

The Steel Diagram

That part of the Iron-Carbon Diagram up to 1·7 per cent of carbon which is below the solidus is called the Steel Diagram. A little examination will show that iron-carbon alloys from 0 per cent to 1·7 per cent carbon begin freezing at the liquidus line (running from 1500° C. to 1360° C.) and finish freezing at the solidus line (running from 1500° C. to 1130° C.). As a rule, engineers will not want to melt and cast steel, but they should know that steels in solid form consist of a solid solution of carbon in iron; that the kind of iron at these high temperatures in the solid state is of an "allotropic" form called gamma (γ); that, on the way down to room temperature (ordinary temperatures) the gamma form changes to an alpha form (α) which cannot retain carbon in solid solution, so that the carbon is precipitated not as carbon but as carbide of iron, Fe_3C (6·7 per cent carbon).

These changes in the iron structure (from the face-centred cubic lattice of gamma to the body-centred cubic lattice of alpha iron) take place in the completely solid state; they can be followed by the cooling curve arrests, and plotted in the steel diagram; their manipulation by heat treatment ought to be understood because they closely control the properties of the steels.

CHAPTER XI

THE STRAIGHT CARBON STEEL DIAGRAM

THE steel diagram, of changes in the solid state between 0 per cent and 1·7 per cent of carbon, has just the shape and general appearance of a diagram of any alloy of A and B. Is this a curious coincidence or not?

As a liquid cools its molecules are coming nearer to one another until at some temperature a new condensation of structure is ready to take place; it takes place when freezing occurs; and, in that condensed structure, the solubility of any solute is often so reduced that complete ejection of the solute occurs. Sometimes the reduction of solubility is far from complete when cases of limited solid solubility occur; sometimes there is no reduction, as in the case of alloys in which there is an unbroken range of solid solutions.

In the case of solid carbon steels, as also in the case of certain other solid alloys, the change from one crystalline allotropic form to another at a lower temperature is also an example of a condensation, usually restricting the solubility of any solute which the former structure may have harboured within its lattice.

Hence there is no coincidence in the similarity between the shape of the steel diagram and the shape of the liquid-solid equilibrium diagram; the causes of each are identical.

The presence of a solute in a liquid metal commonly has the effect of reducing the temperature at which freezing commences, causing the act of freezing to occur over a range of temperatures instead of at one specific temperature; exactly the same phenomena accompany the changes in the solvent structure when it is a solid solvent like gamma iron. This is not a liquid which freezes, but a solid which reaches a "change point"; the change point is lowered more and more by greater and greater amounts of carbon solute in it, down to a minimum, after which the change point rises again as the other side of the diagram is approached—as in the case of the alloy of A and B.

For pure iron (0 per cent carbon) the gamma to alpha condensation or allotropic change point or "critical point" (also called the "recalescence point") occurs at 900° C.; by successive additions of carbon up to 0·9 per cent the critical point is reduced to 700° C.: further additions of carbon up to 1·7 per cent cause the critical point to climb from 700° C. to 1130° C.

What actually happens, then, on cooling steels down? What is the observable effect?

A "hypo-eutectoid" steel, that is, a steel containing less than 0·9 per cent of carbon, will remain just a gamma solution of carbon (resembling pure iron under the microscope) until it is cooled to its "upper" critical point; then a little of this iron will change to the alpha variety which does not dissolve carbon at all, hence its carbon will increase the carbon contents of the remaining gamma iron, so that the change point of the latter will drop to a lower temperature; in due course the temperature will fall to this lower change point, and more gamma will change to alpha, throwing its carbon in turn on to the yet remaining gamma, and still further reducing the change point temperature. These steps downward will be so small that the effect will be a sliding down the sloping line of the diagram.

Therefore, the gamma will be changed to alpha over a range of temperature down to 700° C., by which time the strength of the remaining gamma will be 0·9 per cent carbon.

So that at 700° C. the mass of steel consists of a certain amount of alpha iron and a certain amount of residual gamma solution of 0·9 carbon strength; the quantities of each can accurately be predicted from the diagram, and these depend entirely upon the carbon percentage of the original steel. The next thing to happen is the formation at this temperature *from the residual gamma* of a eutectoid of the components on either side of this steel diagram, a eutectoid of alpha iron and carbide of iron, Fe_3C. 700° C. is the temperature of the "lower" critical point.

The alpha iron occurring as above either free or in the eutectoid is called "*ferrite*"; the eutectoid as a whole is commonly called "*pearlite*."

This is the first time the component on the right-hand side of the diagram has been mentioned: it is this carbide of iron, called "*cementite*," which is the "primary" to the right of the steel diagram, and the mention of it now brings the opportunity of explaining what happens when a hyper-eutectoid steel, one containing more than 0·9 per cent carbon, is cooled down from the gamma state of its iron.

No change of any kind takes place until such a steel reaches the upper critical point shown in the diagram, and then Fe_3C begins to separate out of the gamma solution; this separation of a carbon-rich compound *weakens* the carbon content of the remaining gamma, so that the change point of the latter drops, and is only overtaken by the cooling a second later, when more Fe_3C separates, causing

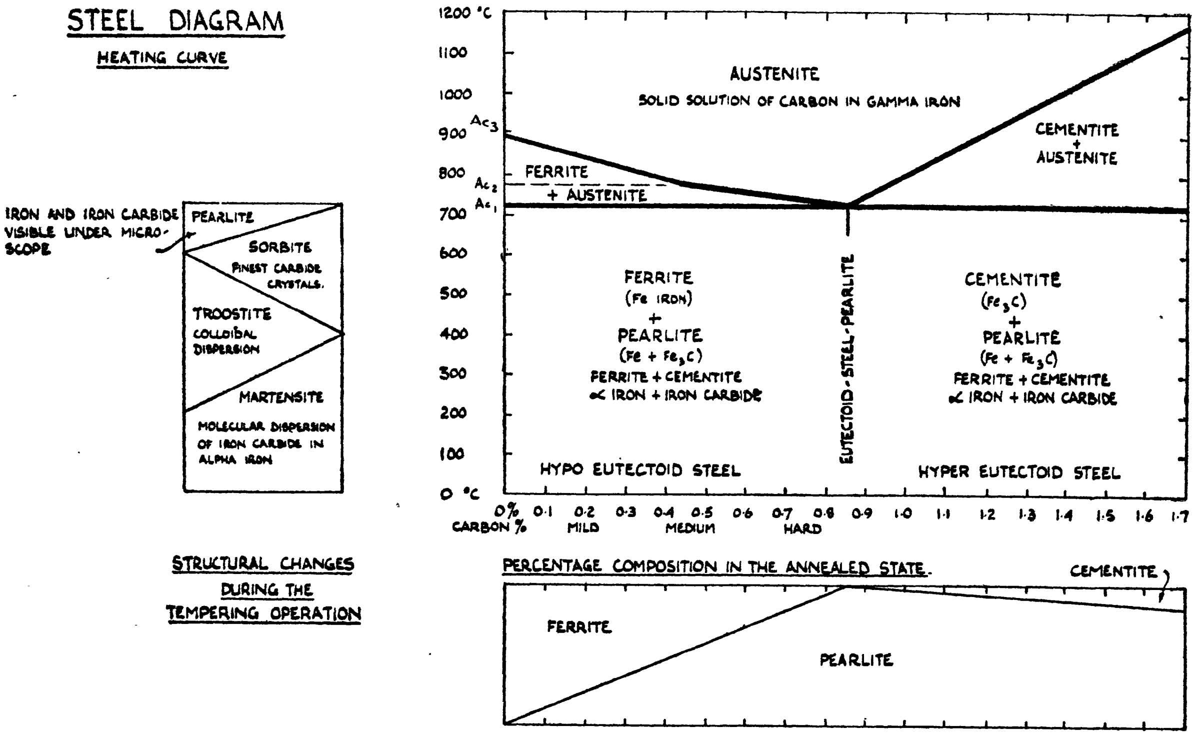

STEEL DIAGRAM
HEATING CURVE
IRON AND IRON CARBIDE VISIBLE UNDER MICROSCOPE
PEARLITE
SORBITE
FINEST CARBIDE CRYSTALS.
TROOSTITE
COLLOIDAL DISPERSION
MARTENSITE
MOLECULAR DISPERSION OF IRON CARBIDE IN ALPHA IRON
STRUCTURAL CHANGES DURING THE TEMPERING OPERATION
1200 °C
1100
1000
900
800
700
600
500
400
300
200
100
0 °C
Ac3
Ac2
Ac1
AUSTENITE
SOLID SOLUTION OF CARBON IN GAMMA IRON
FERRITE + AUSTENITE
CEMENTITE + AUSTENITE
FERRITE (Fe IRON) + PEARLITE (Fe + Fe3C)
FERRITE + CEMENTITE
α IRON + IRON CARBIDE
EUTECTOID - STEEL - PEARLITE
CEMENTITE (Fe3C) + PEARLITE (Fe + Fe3C)
FERRITE + CEMENTITE
α IRON + IRON CARBIDE
HYPO EUTECTOID STEEL
HYPER EUTECTOID STEEL
0% CARBON %
0·1
0·2
MILD
0·3
0·4
0·5
MEDIUM
0·6
0·7
HARD
0·8
0·9
1·0
1·1
1·2
1·3
1·4
1·5
1·6
1·7
PERCENTAGE COMPOSITION IN THE ANNEALED STATE.
CEMENTITE
FERRITE
PEARLITE

Fig. 16

a further drop in change point, and so on until the residual gamma solution is at the eutectoid composition 0·9 per cent carbon at 700° C. Finally, at 700° C., the eutectoid forms as in the case of the hypo-eutectoid steel.

If a steel starts with a composition of 0·9 per cent carbon it is said to be the "eutectoid steel"; it has only one change point into eutectoid at 700° C.

Hypo-eutectoid Steels

Hypo-eutectoid steels, therefore, have a microscopic structure of pearlite, or eutectoid (more or less, according to the carbon contents) in ferrite; and the eutectoid structure, like the eutectic structure, is a mixture of alpha iron and carbide visible under the microscope.

But this steel eutectoid is a mixture in a form extremely fortunate from the standpoint of mechanical properties; not only is it very fine, but it consists of fine *layers or laminae* of the components, one of which is soft ductile iron and the other hard and brittle carbide; throughout any particular grain of this eutectoid the laminae are contorted so that there is no direction or plane of weakness anywhere in it; it is a structure (easily visible) of unusual toughness for a eutectic or a eutectoid.

Here, then, is a relatively coarse structure (in the sense that it is very far from being sub-microscopic) which does have certain good effects on properties; its presence in ferrite raises elastic limit, yield point, strength, and hardness, with a certain proportionate decrease in ductility and loss of toughness. The tendency to the sacrifice of ductility and toughness as strength and hardness are increased is very marked in the case of metallic materials; but the whole aim of the metallurgist is to maintain the former at the necessary high level whilst the latter are improved; and the development of the pearlite structure in iron with the aid of carbon up to 0·9 per cent is a definite move in the right direction so far as it goes.

Hyper-eutectoid Steels

Hyper-eutectoid steels contain "primary" carbide (better called "free," or "pro-eutectoid" carbide) which, in itself, is very hard and brittle; but this constituent is so rich in carbon (6·7 per cent) that a little of it goes a long way to supplying the extra carbon which makes the steel exceed the eutectoid composition, and so the amount of the free carbide in steels in common use (up to 1·2 per cent carbon) is not very great. This free carbide gives a coarse-grained (annealed) steel of diminished strength and increased ductility; in a fine-grained state the free carbide has the effect of further increasing the elastic limit, strength, and hardness, and in decreasing the ductility.

Hysteresis

Fig. 17 will repay careful study. The "r" comes from "refroidissement" (cooling), and the "c" from "chauffage" (heating). At *practicable* cooling and heating rates the true change points are always overshot.

Optimum Properties

It has already been seen that fine grain by "normalizing" (or refining) effects improvement; but by utilizing a special treatment of quenching and tempering still finer invisible structures will be found to confer "optimum" properties—best ductility and notch-toughness for a given strength.

Quenching and Tempering

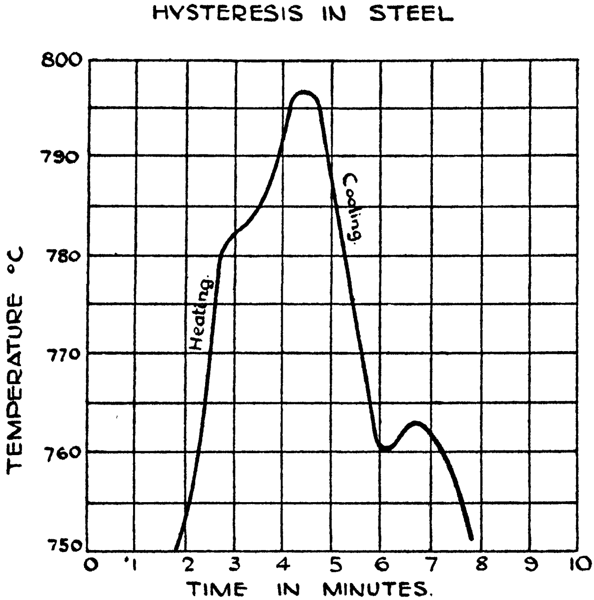
Treatments whereby the grain size and the structure visible within the grain are modified can both be viewed as treatments to change the microscopic structure.

But changes in the sub-microscopic structure, in the lattice structure, seem to be more far-reaching in their effects; a means has been discovered whereby the sub-microscopic structure of carbon steels can be modified so as to provide *a series of properties* for steels of *each* carbon percentage, among which are to be found the real "optimum" properties for any straight carbon steel.

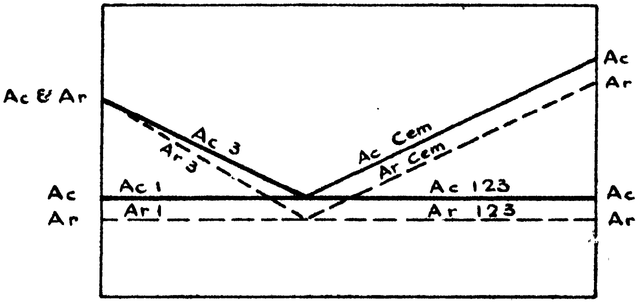
At temperatures above the upper critical points, steels consist of a homogeneous solid solution of carbon in gamma iron; this has received a special name, and will in future be called "*austenite.*"

By extremely quick cooling from a higher temperature it is frequently possible to trap an alloy structure at that temperature, preventing the change which would normally take place under very slow (equilibrium) conditions; but in the case of the carbon steels, it is well-nigh impossible to trap austenite by quenching. What is trapped in this way is a structure well known as "*martensite,*" which is produced as a result of *the first step towards* the formation of the proper equilibrium structure: however fast the austenite of carbon steels is cooled it is found impossible to catch more than a negligible quantity of the austenitic structure: it changes so fast to the first "transition product"—to this *martensite* (see Fig. 18).

The very first step which the austenite must take before becoming equilibrium steel, or normal steel, is to change its gamma iron to alpha iron. Now, this form of iron has no power of dissolving carbon; if the change is instantaneous (as it seems to be) at some


HYSTERESIS IN STEEL

HEATING FOLLOWED BY COOLING IN A CONTINUOUS TIME TEMPERATURE CURVE – SHOWING A HYSTERESIS OF MORE THAN 20°C AT THE SAME CHANGE POINT.



Ac STANDS FOR HEATING POINTS, e.g., Ac 1., Ac 3.
Ar STANDS FOR COOLING POINTS, e.g., Ar 1, Ar 3

THE STEEL DIAGRAMS BY HEATING AND COOLING.

FIG. 17

temperature on quenching, then the alpha may be *choked or supersaturated* with the carbon for an instant; but, almost as soon, the bulky molecules of carbide of iron *form inside the lattice structure of the alpha.*

Signs of a partially successful attempt on the part of this carbide to precipitate can be seen under the microscope in a *martensitic microscopic structure,** but the properties of steels in the martensitic state are due to the sub-microscopic first appearance of the bulky carbide molecules in the alpha lattice structure whereby enormous lattice pegging renders the steel excessively strong, hard, and brittle.

The properties of martensite, the extraordinary accession of strength, hardness and brittleness derived from the soft ductile austenite by quenching, call for the formulation of correct ideas, not only because the quench-hardening of steels is important, but because similar effects occur in the case of many other alloys, all due to precisely similar causes. If the wrong argument is adopted to account for the quench-hardening of steels, it is utterly impossible to co-ordinate, or to make head or tail of, the effects in the other alloys.

Quench Hardening

Quench hardening, or "hardening right out," is therefore produced in any carbon steel by quenching the steel from above its upper critical point. It is advisable to quench *from only just above,* because otherwise the advantage of smallest possible grain-size will be lost.

Tempering

Tempering is the further heating, after the quench hardening, whereby any desired amount of ductility (and toughness) *may be restored* to the steel. Taking the martensitic structure up to higher and higher temperatures is said to "let down" the hardness and brittleness of the steel, to temper the prior properties of hardness and brittleness. Tempering temperatures must all be below the critical range, however.

Upon so doing, the first effect observed under the microscope is the change from the characteristic extremely fine needle structure of martensite to a nebulous nodular structure difficult to identify without much experience. This is known as "*troostite.*"

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* The visible (Widmanstätten) structure here may, in reality, be due more to the gamma to alpha change than to carbide precipitation. (Compare al. bronze.)

Tempering at a higher temperature will again change the microscopic appearance of the steel, producing a structure known as "*sorbite.*" Even this is difficult to identify without considerable experience, although there is no question about the ease with which ideal examples of troostite and sorbite can be distinguished from each other or from martensite. Ideal sorbite will, under highest magnifications, look like tiniest specks of black uniformly and very closely packed in a white matrix, but it may also resemble an excessively fine dense eutectoid of steel (see Fig. 19).

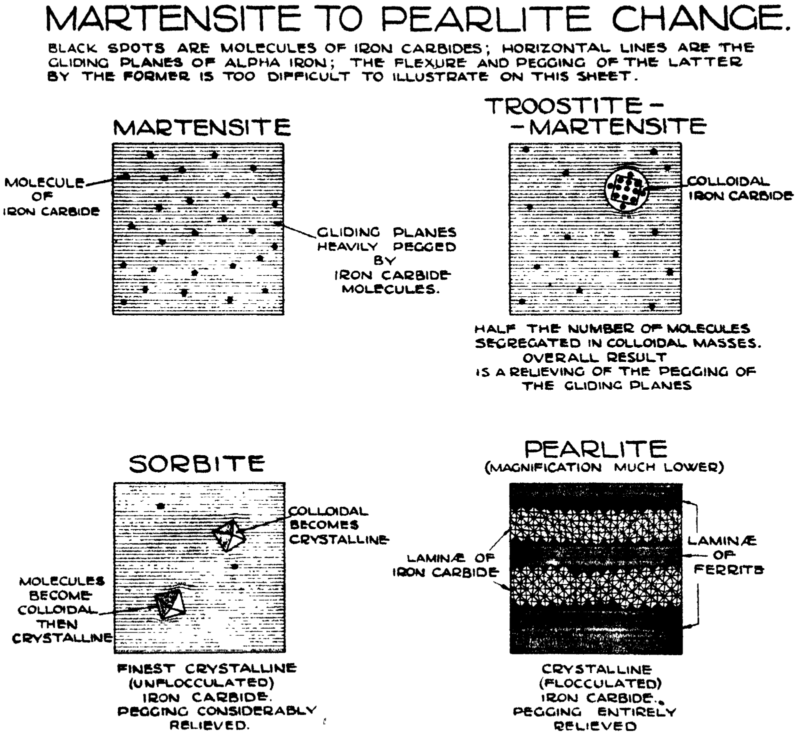

Fig. 18

Sorbite is, in fact, the last transition product between austenite and the equilibrium or normal structure* of steel (ferrite-eutectoid, eutectoid, or cementite-eutectoid); a type of sorbite known as lamella sorbite resembles a very fine and uniform eutectoid structure; the type known as granular sorbite is just a fine close speckling in black on a white background.

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* See last paragraph on page 121.

THE AUSTENITE TO PEARLITE CHANGE THROUGH TRANSITION PRODUCTS.

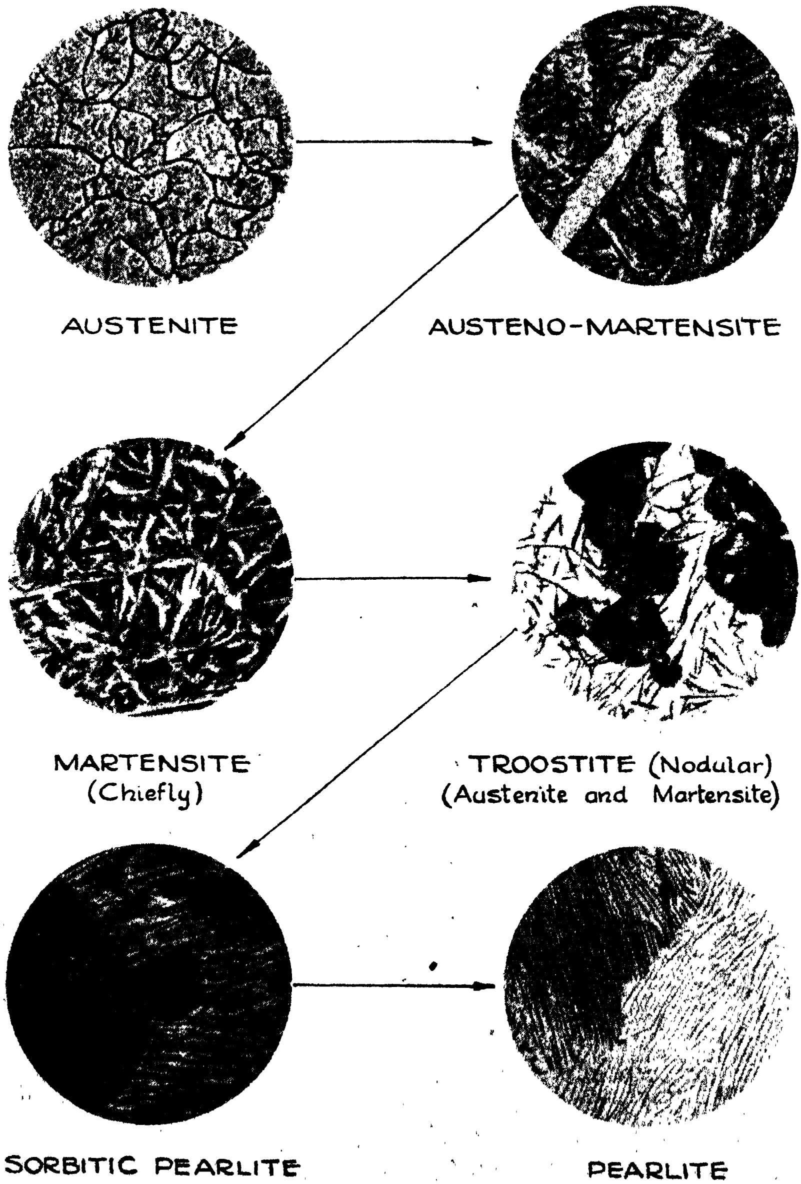


Fig. 19

Some important points remain to be touched upon—

(i) These structures are not peculiar to steel; but they are only the outward and visible signs of sub-microscopic transitions from one equilibrium state to another, and are *found in other alloys* similarly caught in acts of transition.

(ii) These structural arrangements should not be looked upon as separate alloys, for they are only different appearances in the alloys known as steels (bronzes, etc.).

(iii) It should not be assumed that steel with 100 per cent martensitic, troostitic, or sorbitic structure, can be obtained in practical tempering, as there will rarely be clear-cut differences, but only a gradual passage from austenite to pearlite steels, in which passage the three "transition products" are fairly easily recognizable steps, stages resolvable by the microscope. But it is nevertheless correct and essential to regard martensite as a steel in that section of the transition range where the carbide is substantially as a *molecular dispersion*; troostite as a steel in that section where carbide is substantially as a *colloidal dispersion*; and sorbite as a steel in that section where the carbide is substantially as *finest crystalline* (but not yet flocculated into lamellae or granules) dispersion.

Hence, each of these transition products can and should mean a *definite "metallic structure"* responsible for the mechanical properties of the product.

(iv) The whole point of this heat treatment is that, by adopting it, better toughness and notch toughness can be obtained for any required strength and hardness; also, a range of properties varying from soft, weak, and tough, to hard, strong, and brittle, can be obtained *in a steel of each carbon percentage.*

(v) Substantially pure martensite, troostite and sorbite may also be obtained by exceptional methods of cooling which need hardly be discussed because they cannot be employed in regular practice. (See "Austempering.")

(vi) The tempering of quench-hardened steels at temperatures from 600° to 650° C. (higher than those to produce sorbite) gives rise to carbide dispersions of a degree coarser than sorbite in a spheroidal rather than a laminated form. The form of pearlite thus obtained therefore shows no pearly lustre, and is sometimes called "granular pearlite."

CHAPTER XII

THE PROPERTIES OF STRAIGHT CARBON STEELS

PURE iron is no exception to the general rule with metals that traces of certain common impurities affect properties; so that neither wrought iron nor the much purer "Armco" resemble perfectly pure iron very closely. The influence of such traces is often felt in the grain boundaries, causing some weakness where metals can be strongest; often, also, there is a little lattice strengthening, etc.

Carbon steels (containing no more than the usual permissible maxima of silicon, manganese, sulphur and phosphorus) exhibit changes in properties, due solely to changes in carbon content, approximately as follows—

The Effect of Carbon Percentage

All properties vary *linearly* from 0·1 per cent C. (nearly 100 per cent ferrite) to 0·9 per cent C. (nearly 100 per cent pearlite) after which free cementite makes its appearance, and the linear relationship is not so marked.

Between ferrite and pearlite *in the annealed state*, as U.T.S. rises from 18 to 36 tons sq. in., ductility falls from 50 to 5 per cent elongation; *in the normalized state*, as U.T.S. rises from 22 to 65 tons sq. in., ductility falls from 35 to 5 per cent elongation. There are bound to be variations, however, with changes in the amounts of the various impurities. Normalizing may improve elastic limit, yield point and notch toughness more than U.T.S., and hardness in normalized steel will vary throughout the range of carbon from 100 to 300 B.H.N. whilst reduction of area at fracture will vary from 65 to 10 per cent.

In heat treatment for optimum properties by quench-hardening and tempering, an infinite number of compromises between certain limits of U.T.S. and ductility may theoretically be obtained for each carbon percentage by variations of the tempering temperature. The effect of such heat treatment actually varies from practically nothing in the mildest steels to an important maximum in the case of the high carbon steels; but the increases in the effect with increasing carbon content are not linear but much greater for the initial increases in carbon and subsequently less for later increases in carbon. Hence quench-hardened and tempered medium carbon steels are to be preferred for many purposes to steels of equal strength in normalized higher carbon steels.

A 1·2 per cent C. steel may have its U.T.S. varied by tempering from 100 to 50 tons sq. in., whilst ductility varies from 2 to 15 per cent elongation, and hardness from 430 to 220 B.H.N.; the fatigue limits of the strongest of these are very high (48 tons sq. in.), though their notch toughnesses are very low indeed.

The Effect of Different Treatments

Upon a steel of one carbon percentage (0·35 per cent) the effects of the three common heat treatments can be seen in the following table. (The tempering temperature in the treatment for optimum properties has been chosen to procure a ductility comparable with that of the other treatments).

	Proof Stress	Y.P.	U.T.S.	% El.	% R.A.	Fat. Limit	B.H.N.
(1) Annealing . .	17	18	36	24	39	14	157
(2) Normalizing . .	19	20	37	23	36	15	164
(3) Quench-hardening and tempering .	23	27	43	26	49	16	190

The notch toughness values of treatments (2) and (3) are also much better than in the case of the annealing treatment.

The Effect of Different Tempering Temperatures in the third treatment is quite commonly shown on charts made for different carbon percentages. Between tempering temperatures of 200° C. and 700° C. all properties change linearly from those of martensite (which is strong, hard, etc., but very inductile and notch sensitive) to those of the pearlitic structure; and, by taking tempering (followed by air cooling) still higher to the original quench-hardening temperature, it is of course possible to attain the normalized properties again, because the quench-hardening and the normalizing temperature is the same (just above the A_c3 point).

As already mentioned, the effect of quench-hardening on the mildest steels is negligible, meaning that the martensitic properties of these do not differ appreciably from their pearlitic or normalized properties; though, on proceeding from low to medium and high carbon content, martensitic properties very soon come to differ enormously from the pearlitic and the normalized properties.

Optimum Properties of Practical Value may therefore be obtained by tempering between 250° C. and 650° C. At 250° C. reasonable ductility and notch toughness have appeared with very high strength, hardness, etc. At 650° C., with full ductility and notch toughness, strength, hardness, etc., are still definitely higher than can be obtained by any other method of treatment giving the same ductility.

The Effect of Cold Work is roughly exemplified in the drafting of a 0·6 per cent C. steel. U.T.S. can be said to increase almost linearly—up to a drafting reduction of 70 per cent—from 60 to 100 tons sq. in.; whilst at the same time ductility, dropping rapidly from 12 to 3 per cent elongation, remains at about 3 per cent for the last two-thirds of such a drafting range. For further drafting up to 95 per cent, U.T.S. rapidly increases up to 170 tons sq. in. whilst percentage elongation drops off to 1 per cent. Of such heavily cold worked steels of this composition, the contraction of area at fracture is dropping rapidly from an average of 45 per cent in the earlier stages to zero at 100 per cent cold work.

The Effect of High Temperature on the properties of iron and steel is, in general, to lower the limit of proportionality, elastic limit and yield stress even from the first temperature increases. In the case of the U.T.S. it is a different story, for strain ageing is increased to such an extent by slightly increased temperature that after a little fall as far as 100° C. or so, U.T.S. goes higher again up to a maximum at about 300° C., at which an increase of 5 tons sq. in. or more may be registered in the tensile tests of iron and steels. From this highest figure of U.T.S. there is a steady falling away with increase in temperature beyond 350° C. until, at 700° C., all carbon steels reach U.T.S. values of from 5 to 10 tons sq. in. according to carbon contents.

Both percentage elongation and contraction of area follow the above U.T.S. changes by dropping at the U.T.S. maxima, and then by increasing fairly regularly with increase in testing temperature until, at 700° C., they are approximately 70 and 95 per cent respectively. Hence it is dangerous to try to hot work iron and steels at this "black heat" temperature of drop in ductility.

The Heat Treatment of Steels

Taking the widest view of "heat treatment," a steel with a satisfactory chemical analysis might be utterly unfitted for use on account of defects of the structures upon which its properties depend so closely.

The following applies to any carbon steel—

Rate of Cooling

In the absence of any easily applied standard for cooling rate a certain amount of confusion of thought often arises: it is usual to accept cooling in the furnace as the standard slow rate of cooling, cooling in air as the standard medium rate, and quenching in water as the standard fast rate. For small hand-specimens this classification holds good, but the mass-surface ratio is an important factor.

Every practical case depends upon the mass and surface of the steel being cooled, so that the cooling of a knife blade blank in air will be fast cooling compared with that of a crankshaft in air; some large masses are cooling at a slow rate even when they are being sprayed with water, so that special or alloy steels must be used for the special purpose of hardening the interior of large masses.

Burnt Steel

A rod, bar, sheet, or any other cast or machine-fashioned piece of steel may have been exposed to very high temperatures in air. The usual heavy coating of scale might have been removed; and, if so, only the microscope would show in advance that such a specimen would have very poor properties; it should be possible to discover signs of oxidation in the grain boundaries extending from the exterior into the mass. It might be possible to observe also that the grain size is coarse, though grain refining might have been carried out after burning.

Now, coarse grain alone causes some notch brittleness in comparison with a fine grain condition: but these oxidized boundaries give rise to a useless condition of brittleness from fractures starting at such hair cracks in the metal's surface.

The only way to recover such material for use is to remelt altogether, or else to remove the oxidized exterior and to forge the remainder at welding heat (near melting point)—subsequently "refining" according to the next treatment.

Overheated Steel

In this case the steel has been taken to and kept at high temperatures (far over its A_c3 point) for too long a period of time: a higher temperature for a short time or a lower temperature (well over the A_c3 point) for a long time will permit "grain growth" whether the atmosphere be oxidizing, reducing, or neutral.

Overheated steel is not oxidized: there are no burnt grain boundaries as in the case of burnt steel, so the remedy is—

"Refining" (sometimes called "Normalizing," particularly when an identical treatment is used for relieving stresses). This consists in reheating the steel to 10 degrees or so above the A_c3 point, and then cooling in air (medium cooling rate).

The cause of this "refining" of grain has already been given in Chapter V under the heading "Metastable and Labile States." On heating pearlite through its A_c1 point, change to austenite takes place from a number of different points in each pearlite grain, thus producing more austenite grains than the original number of pearlite grains. On cooling these smaller austenite grains through the A_r1 again, pearlite forms from a number of different points in each austenite grain.

Very much more refining takes place during the heating than during the cooling for some reasons which are not absolutely clear.

Steel Suffering from Internal Stresses

Steel suffers from internal stresses as a result of casting in which the outside has cooled at a faster rate than the inside, or as a result of quenching, or of cold working (considering the residual elastic stresses, not the cold work hardening effect).

The danger here is that new stresses developed as loads are applied, in themselves safe, may be unsafe when added to some resident internal stress; fracture may then take place at uncommonly low apparent stresses.

Annealing or Normalizing heat treatment may be used to remedy this defect.

Annealing

Annealing aims at softening the steel for further work in the cold; the treatment merely consists in heating to a temperature 20 degrees above the A_c3 point and cooling in the furnace.

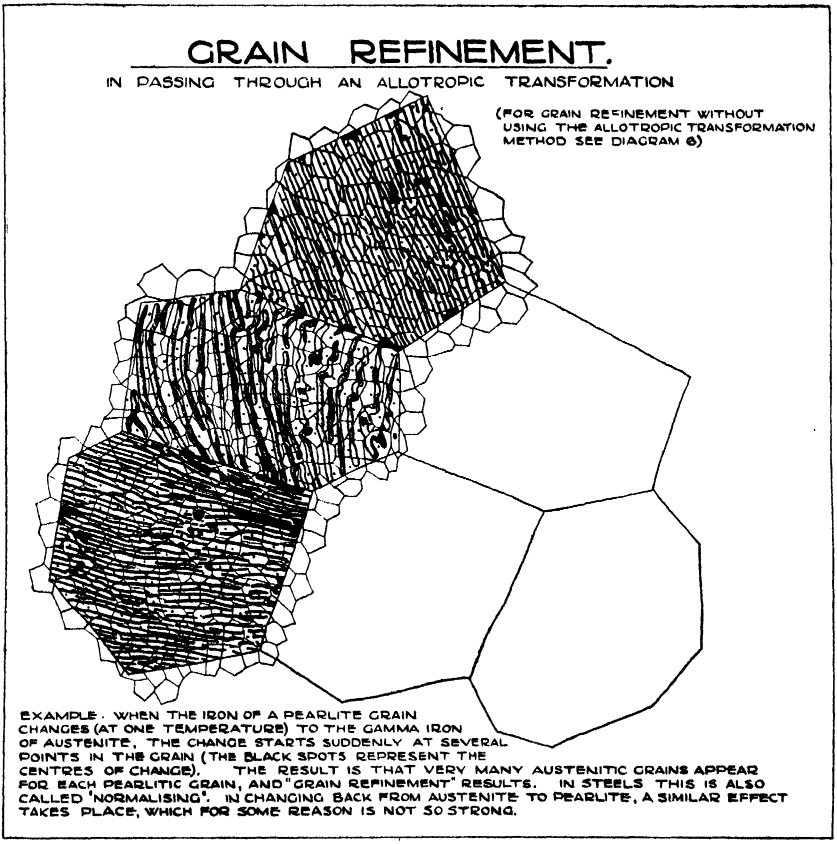
Normalizing

Normalizing aims at endowing the steel with superior properties for use—not so suitable for further machining, etc.: the treatment is as for Refining.

Stresses may be removed by heating to temperatures short of the A_c3 point, but such heat treatment would take longer and might neither soften as in annealing, nor give superior properties for actual use as in normalizing. (Compare Spheroidizing.)

Steel Suffering from Strain Hardening

Plastic deformation results in the formation of hardening amorphous cement on the slip bands of the metal grains which may not be desired. It may be found necessary to reduce or remove this strain hardening and to regain a softer and a more ductile metal.



GRAIN REFINEMENT.

IN PASSING THROUGH AN ALLOTROPIC TRANSFORMATION

(FOR GRAIN REFINEMENT WITHOUT USING THE ALLOTROPIC TRANSFORMATION METHOD SEE DIAGRAM 6)

EXAMPLE. WHEN THE IRON OF A PEARLITE GRAIN CHANGES (AT ONE TEMPERATURE) TO THE GAMMA IRON OF AUSTENITE, THE CHANGE STARTS SUDDENLY AT SEVERAL POINTS IN THE GRAIN (THE BLACK SPOTS REPRESENT THE CENTRES OF CHANGE). THE RESULT IS THAT VERY MANY AUSTENITIC GRAINS APPEAR FOR EACH PEARLITIC GRAIN, AND "GRAIN REFINEMENT" RESULTS. IN STEELS THIS IS ALSO CALLED 'NORMALISING'. IN CHANGING BACK FROM AUSTENITE TO PEARLITE, A SIMILAR EFFECT TAKES PLACE, WHICH FOR SOME REASON IS NOT SO STRONG.

FIG. 20

Complete softening is carried out by annealing at temperatures above the A_c3 point for some time and cooling in the furnace, but sufficient softening may be possible at lower temperatures. For superior properties without strain-hardening, refine or normalize as before.

Steel which is Too Soft

Steel in too soft a condition to use may have been annealed (cooled very slowly) and hence may be sufficiently prepared for many uses by refining (the same treatment as for normalizing).

On the other hand, it may be worth while to take extra trouble to obtain one of the many products of quenching and tempering which are valuable not only because of the alternatives in properties so amply provided, but because maximum toughness is provided with any desired strength and hardness. If so, quenching and tempering are resorted to as follows—

Hardening "right out" is effected by quenching in water from just above the A_c3 point; the product is now the hardest, strongest, but also the most brittle possible steel—martensite.

Tempering. Reheating to 200° C., 300° C., 400° C., etc., etc., followed by quenching (or just cooling off in air) "lets down" the original hardness and strength, and induces ductility progressively by allowing the lower transition products (troostite and sorbite) to appear; so that every degree of ductility (and therefore toughness) may be induced at will. Thereby, higher combined strength, ductility, and notch toughness are obtained.

Steel in too hard a condition to machine or to use may not suffer from the amorphous cement of strain hardening, but from the hardening and tempering processes just discussed. If so, the remedy is to replace the martensite-troostite structure, or the troostite-sorbite structure, or the sorbite-pearlite* structure (as the case may be) with a normal coarse pearlite* structure by means of a full anneal.

Spheroidization

For still softer steel the pearlite structure may be "spheroidized" or softened by holding the steel at a temperature just below the A_c1 point: the effect is a segregation or balling-up of the laminae of cementite, a coarsening of the pearlite structure, and a suitable softening for machining.

Forging and Hot Working

Forging and hot working steels generally ensure fine grain and good properties if carried on down to dull red heat, but not below.

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* See last paragraph on page 121.

Cessation of hot working at higher temperatures than dull red permits of grain growth: continuation of hot working at a black heat has a very bad effect as the properties are poor at this low temperature.

" Austempering "

It has recently been discovered that exceptionally good toughness and ductility, together with high strength, may be obtained by quenching carbon steels in a medium at 200° C. to 400° C., and holding for some time within this range of temperature whilst new structures develop. Research on exceptional cooling methods is in progress.

CHAPTER XIII

THE CASE-HARDENING OF STEEL

TOOLS and certain parts of machines may be required to have an extremely hard wearing surface, but this must be supported by a tough backing capable of withstanding sudden stresses, i.e. shock loading. A gudgeon pin must have these qualities.

A little consideration shows that these conditions cannot be obtained with a normal straight carbon steel. The hardness of a quenched 0·9 per cent carbon steel and the toughness (ability to withstand shock) of a 0·15 per cent carbon steel are required. If a piece of 0·15 per cent carbon steel is taken, the carbon content can be increased to 0·9 per cent. for a small depth below the surface. Actually, very near the surface the carbon content will have to be approximately 1 per cent, because it decreases gradually with depth, and it is usual to finish the articles by a final grinding process which removes the outer skin, thus leaving a surface with an approximate 0·9 per cent carbon content.

A reference to the equilibrium (heating) diagram* for carbon steels at once shows that—

(*a*) The minimum solid solution temperature for a 0·15 per cent C. steel is approximately 880° C.

(*b*) The minimum solid solution temperature for a 1·0 per cent C. steel is approximately 720° C.

In actual practice both temperatures would have to be increased by approximately 15 to 20 degrees to ensure that the heat lost when removing the article from the furnace to the quench did not cause the temperature to drop below the A_r3 temperature before immersion.

Given a high carbon case and a low carbon core we can—

(*a*) Heat treat the core for strength by heating the steel to a temperature in the region of 900° C. and then quenching;

(*b*) Then heat treat the case for hardness by heating to a temperature in the region of 740° C., and then quenching. The effect of the 740° C. on the core will be that of tempering, and the core structure will mainly consist of a granular pearlite or

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* Such a diagram gives A_c points; an equilibrium (cooling) diagram shows A_r points which are 30° to 40° lower than the A_c points.

sorbite in alpha iron. In such a condition the core will be tough and able to resist shock loading, but the surrounding case will be hard and therefore wear-resisting.

Preparation of Article

After machining, only sufficient surplus material should be left on the article to ensure that it may finally be trued up to size by grinding. This is essential because final grinding is necessarily a slow process due to the fact that a heavy cut would produce excessive local surface heat, and such heat would tend to soften the skin on spots or patches. In addition a heavy cut causes heat and local expansion which the hardened surface is too rigid to accommodate, and thus cracks or flaking may be started.

Depth of Carburization

Only a small depth of the higher carbon content is necessary, and in most cases would not exceed $\frac{1}{64}$ to $\frac{1}{32}$ of an inch.

As the depth of penetration depends upon the temperature and the time the article is subjected to the high temperature, it is uneconomical to carburize deeper than necessary. The case will also have a different expansion and contraction to the core, and the thinner the case the more easily will it adapt itself to the core.

A very thick case certainly tends to separate from the core during the quenching operations.

The carbon content of the case depends mainly upon the type of carburizing material used.

Carburizing

There are several methods—

METHOD 1. A variety of carbonaceous materials are in general use, including wood or bone charcoal, charred leather, crushed bone, and these are often mixed with an ingredient such as barium carbonate.

It is usual to buy the case-hardening compound already prepared, and it may be obtained from almost any of the factors of engineering sundries.

The article to be treated is packed in the compound in a mild steel box and the top is sealed with fire-clay to keep out air. The box is then placed in a furnace and the temperature raised to approximately 900° C. and left at that temperature for a period depending upon the depth of case required. For a normal depth of, say, $\frac{1}{32}$ in. a period of about $3\frac{1}{2}$ hours would suffice.

The box is then allowed to cool either by allowing the furnace to cool or removing the box from the furnace. The article is then taken out of the box and heat treated as usual.

> NOTE. (*a*) If the article is of an intricate shape it may be quenched in oil instead of water, in order to avoid the possibility of cracking at a change of section, etc.
>
> (*b*) Any portion which it is necessary to retain in the soft condition may be stopped off with either a coating of clay-wash or by copper-plating, prior to packing in the box. (Various preparations are sold for stopping-off purposes.)

METHOD 2. This involves the use of a bath of molten sodium cyanide at a temperature of approximately 900° C. in which the article is immersed for a period of thirty to forty minutes. The articles should be given the same heat treatment as before.

Care must be taken to ensure that the copious fumes given off by the bath are carried away to the outside of the building and are not inhaled by the operator.

This method gives a sound hard case where only a moderate depth is required.

METHOD 3. This is an interesting method which seems to work well. A mixture of salts, supplied under a trade name, "Perliton" (melting point 593° C.), is heated to a temperature of 900° C. in an ordinary pot salt bath.

A special carbonaceous substance is then added slowly (about 5 lb. for, say, 100 lb. salt) and the bulk of this goes into solution, the remainder floats on top of the salt. The article to be carburized is immersed in the bath for a period of one to four hours and the depth of penetration is 0·02 in. to 0·04 in. according to time.

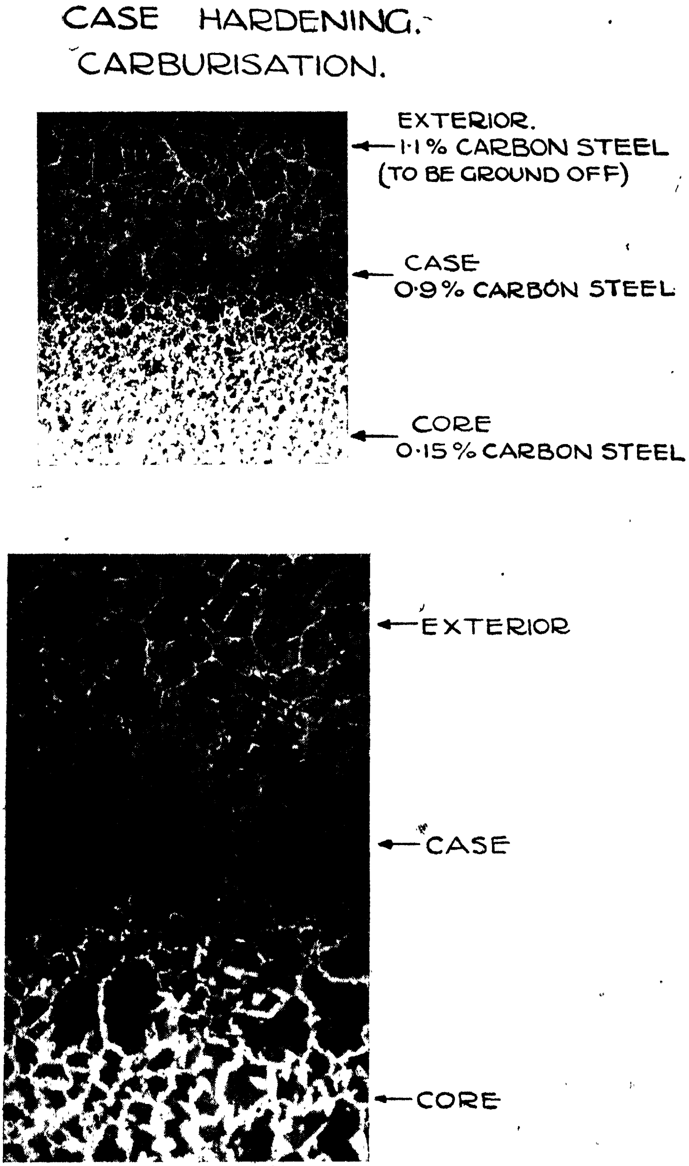
Quenching may be carried out straight from the salt bath and the subsequent heat treatment is as usual.

This method is a good solution of the case-hardening problem, and the following advantages can be claimed for it—

(i) It is very economical as regards time taken for the whole process.

(ii) It gives a case of about 0·9 per cent to 1 per cent carbon and a surface free from defects.

(iii) The article is quite clean and free from scaling after treatment.



AS ABOVE
(HIGHER MAGNIFICATION.)

FIG. 21

(iv) The bath is free from objectionable fumes.

(v) The depth of penetration is easily controlled by the time factor.

Nitralloy Steels

The nitriding process for producing an extremely hard surface on certain steels has only been in commercial use for a comparatively short time.

The hardness that can be obtained when using suitable steels is of the order of 1000 Diamond, which is greater than the hardness normally attained by the carburizing process.

Straight carbon steels are not greatly affected by the process, the gain in surface hardness being comparatively small. The addition of nickel does not increase the effect. The essential alloying elements are aluminium, chromium, molybdenum, vanadium. Aluminium is not used if a very fine finish is required because it has a tendency to make the steel flaky and rough.

The carbon content varies between 0·15 per cent and 0·5 per cent according to the tensile strength required.

In the lower carbon contents the steel can be machined in the suitably heat-treated condition and is usually supplied by the makers in that condition.

To facilitate machining, the higher carbon steels are supplied in the soft condition and must be heat treated as a final operation before nitriding.

All heat treatment and finishing processes such as grinding are completed before nitriding, and owing to the low temperature at which the process is carried out there is complete freedom from scaling or distortion, and dimensional growth is practically negligible.

The Nitriding Process

The articles to be surface hardened are finished to final dimensions. They are then placed in a rectangular sheet metal box and so spaced as to leave room for free circulation of ammonia gas around them. The gas-tight lid of the box is then fixed in position. The box, which slides on rails, is pushed into an electrically-heated furnace and the front of the box forms the door of the furnace. The door is fitted with an inlet and an outlet connection to which are attached pipes for circulating ammonia gas freely through the box. (The strength of the ammonia gas is maintained constant.) The furnace is then raised to 500° C.

The process is continued for a period of eighty hours after which the box is allowed to cool down in the furnace with the gas still circulating. When cold the box is withdrawn and unpacked.

A slight greyish bloom will be noticeable on the surface of the articles, but this is in no way detrimental.

Depth of Hardness

The depth of hardness will be much the same as in ordinary case hardening, i.e. about 0·02 in., but the change from hardened part to unhardened core is very much more gradual and indefinite.

The depth of hardening could be increased (if it was necessary to do so) by increasing the time, but any alteration of temperature is definitely detrimental to the degree of hardness obtained. Nitrogen alone will not harden the steel: only ammonia will do.

Action of Process on the Steel

When aluminium is present it is probably in solution in the ferrite. When chromium and/or vanadium are present they will be partly in the form of carbides and partly in solution in the ferrite.

It seems that the nitrogen in the ammonia gas at 500° C. forms nitrides of aluminium, chromium, etc., according to the elements present.

It would seem that the units of space lattice formed by these elements are converted into nitride units (molecules of compound with nitrogen) which, being so much more bulky, exert much greater pegging in the space lattice, and by so doing increase enormously the resistance on the gliding planes and induce hardness.

Uses of Nitrided Steels

Steels treated by this process are quite successfully used for such purposes as aero-engine crankshafts.

The stems of aero-engine valves made from austenitic stainless steels are very successfully treated by this process.

It should be noted that nitrided steels, like all other surface hardened steels, are very unsuitable for articles subjected to percussion. The hard surface is liable under such conditions to crack and flake.

It is also unsuitable for drawing dies for rods or tubes, having a tendency to become roughened under such severe use.

CHAPTER XIV

INTRODUCTION TO ALLOY STEELS

IN the straight carbon steels hitherto discussed, the alpha iron or "ferrite" constituent, and the ferrite component of the eutectoid may not be quite pure; silicon and manganese commonly occur in the pig iron from which steels are made; manganese is sometimes added intentionally in small quantities to deal with sulphur; phosphorus occurs; titanium may be present: but the proportions of these impurities must be kept low.

Excepting sulphur (which is present in these steels as tiny particles of manganese sulphide uniformly distributed through the mass), all these other small quantities of impurities are hidden away dissolved in solid solution in the iron—in the free ferrite and in the ferrite of the eutectoid also.

These impurities do not greatly modify the properties of pure iron; but if any of them, or if any other element were to be dissolved in quantity, then the mechanical properties of the alpha and gamma iron, and therefore of the steels so produced, would be modified, possibly to a profound degree.

It will be seen that not fundamental but highly valuable changes can be effected in steels by the introduction of special alloying elements in small and medium amounts; in large amounts fundamental changes may give products which cannot be regarded as "steels" at all.

Alloy Steels: Solid Solution and Carbide Formation

All the elements usefully employed in the preparation of hundreds of special steels are capable of going into solid solution in the iron to form special ferrites, but not all of them compete with the iron for the carbon which is present.

Nickel, copper and silicon form no carbides: manganese, chromium, vanadium, molybdenum, and tungsten, do compete with the iron for the carbon; and, as a result, compound carbides or mixed carbides appear which are usually tougher than the cementite of straight carbon steels. And so, wherever carbides are formed, these are usually better than plain cementite.

Solid solutions are always stronger and harder than the solvent metal, no matter what the solute metal or non-metal may be; but where extra strength and hardness are required without undue loss of ductility, then the choice of the solute, the alloying element, is a matter of greatest importance.

The greater the ordinary toughness and notch toughness retained whilst strength, hardness and elastic limit are being increased, the more valuable the alloy for most structural uses.

Alloy Steel Diagrams

Straight carbon steels on cooling from the austenitic range undergo the initial stages of the change to pearlite with such rapidity that highest speed cooling by quenching is found incapable of trapping and fixing much of the austenitic structure itself. (It will be remembered that the first "transition product," martensite, is the structure caught by quenching—not austenite.)

It has been observed, however, that the greater the amount of carbon or the greater the amount of manganese in straight carbon steels, the more there develops a tendency to the trapping of some austenitic structure on quenching.

This tendency is sufficiently marked by the addition of comparatively small proportions of the alloying elements to enable quantities of high carbon austenite to be trapped by quenching in water. For any given percentage of carbon there is a minimum percentage of alloying element which will give 100 per cent austenitic structure on quenching in water; for any given percentage of alloying element (over a certain amount) there is a minimum percentage of carbon which will yield 100 per cent austenite on quenching.

It should be carefully noted that these "minima" alloy steels are not called "austenitic" steels. On cooling at ordinary rates in air they may be pearlitic, sorbitic, or troostitic and are described as such.

When, however, either the alloying element or the carbon (or both) are present in such large proportions that the steels show martensitic or austenitic structures on cooling *in air at ordinary rates*, then only are they described as "martensitic" or "austenitic" steels, as the case may be.

An important advantage to be gained by alloying other metals or metalloids with iron and carbon lies in the fact that, *by means of different rates of cooling*, the whole range of structures from austenite to pearlite inclusive may be fixed with ease. Austenite is our new structure, unattainable in the straight carbon steels; its valuable properties include perfect homogeneity to resist corrosion, and great toughness, non-magnetic and work-hardening properties.

Another and a vital point about alloy steels is that chosen steels will *harden at slow cooling rates*; hence, the interior of a mass of steel with a high mass-surface ratio (which is bound to cool very slowly) need no longer be soft, weak and ductile if it is required to be hard, strong and inductile. By deliberate choice of steel such an interior with a certain slow cooling rate can be obtained with any desired properties.

The Mechanism of the Property Modification in the Alloy Steels

The technique of alloy steel research, production, and utilization has grown to its present impressive stature from earliest gropings in the dark, through the disillusionments and dim realizations of the twilight, into the light of a day in which we can now state with some confidence how the effects are obtained.

It can now be seen that metallurgists are, in effect, using the methods (iii) and (iv) of Chapter VII in combination.

According to method (iii), the iron lattice is braced by pegging with solid solute atoms of metals like nickel, manganese, or chromium, so that the alloy iron—whether it be the gamma iron of austenite, the alpha iron of the transition products, or the free ferrite or the eutectoid ferrite—is stronger, etc., than that of the straight carbon steel, while it does not suffer in ductility to any undesirable extent. This calls for the most careful choice of alloying element or elements according to the purposes intended; because either high elastic limit, proof stress, yield point or hardness, or certain combinations of them may be far more important considerations than the mere ultimate tensile stress; and these properties are affected variably (albeit probably in the same sense) by the elements available. There are also considerations of notch toughness which call for a fine-grained metal even though the cooling rate must be low; corrosion resistance, fatigue resistance, retention of good properties at high temperatures, and resistance to oxidation and to decarburization at high temperatures, are all considerations in the choice.

According to method (iv): the *compound* needed for this method, iron carbide in the case of the straight carbon steels, is toughened by admixture or combination with a new carbide such as Mn_3C, Cr_3C_2, or WC, so that steels in which the composite carbides are *coarse* (as in pearlite or "cementitic" alloy steels) owe some of their good qualities to the mechanical properties of the massive composite carbides. Even so, it is important to draw attention once again to the point already made before—namely, that, whatever the properties of these composite carbides may be in the crystalline massive state, it is erroneous and seriously misleading to attribute to alloy austenites and to alloy transition products any of the properties of these carbides. For the austenites contain no carbides at all (the latter breaking up on entering the gamma solution), while the three transition products contain but molecular, colloidal, and microcrystalline dispersions of the carbides respectively, none of which can be supposed to exert the intrinsic properties of massive carbides by any stretch of the imagination. (Naturally, this argument excludes vagrant, anomalous, or "cementitic" carbides wherever they may occur in the above-mentioned structures.)

Altering the Steel Diagram

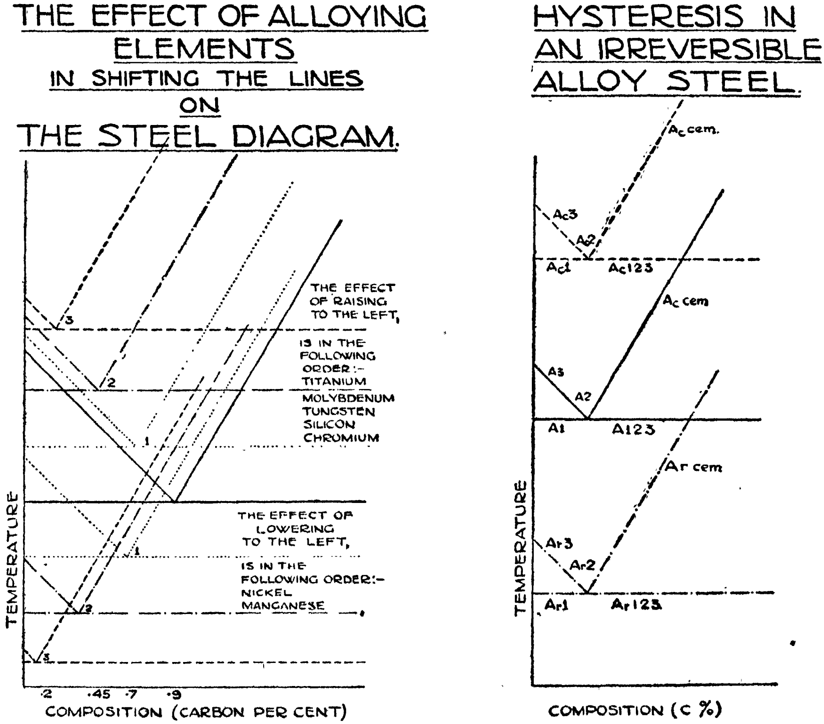
The effects of the alloying elements in producing modification under method (iv) are largely indirect, but are essential and far reaching; for the alloying elements alter the lines of the steel equilibrium diagram. Nickel and manganese pull all the change points down until, with as much as about 25 per cent of either, a steel's austenite is an equilibrium product at ordinary temperatures, and the steels have to be taken to low temperatures to exhibit the transition products or pearlite. Though these particular steels are not in regular use, similar steels of nickel and chromium are in use; but a very common type of alloy steel employs nickel and manganese in sufficient quantity to give sorbite by slow cooling; then there is Hadfield's manganese steel which, on quenching, becomes entirely austenitic.

In fact, these alloying elements have the effect of reducing the rate of cooling required for the formation of any particular structure; thus: cooling more rapidly than in air would be required without nickel or manganese to obtain sorbite, but cooling in air is fast enough if sufficient nickel or manganese be present; and again: austenite would never be obtained by quenching straight carbon steel in water because even this rapid rate of cooling would not be fast enough; but quenching high manganese steel in water will produce completely austenitic structure.

Alternatively, and of equal importance, the centre of a large mass of hot steel must of necessity cool slowly on account of its high mass-surface ratio even if it is dipped in water, or sprayed with it.

Suitable alloying elements like nickel and manganese in correct amounts will ensure that even the slow cooling rate of the interior of this mass is at the right speed to give the structure desired—sorbitic, troostitic, martensitic, or austenitic as the case may be.

Solutes do not invariably lower freezing or change points, and some common alloying elements like chromium and tungsten raise


THE EFFECT OF ALLOYING ELEMENTS IN SHIFTING THE LINES ON THE STEEL DIAGRAM.

HYSTERESIS IN AN IRREVERSIBLE ALLOY STEEL.

FIG. 22

the change points of the equilibrium diagram. But these equilibrium changes call for such infinite patience in cooling and heating that they are impossible to attain in practice; it is commonly found that both chromium and tungsten cause great hysteresis or lag, sending the A_c points high and also dropping the A_r points as well, even though both are raised under strict equilibrium conditions. Use can be made of such a lag, because under these conditions a structure like hard martensite can be obtained in a tool steel at certain high temperatures on cooling and quenching with the knowledge that very much higher temperatures still can be used on reheating before that structure will be destroyed by the resolutioning of the carbides. Such steels are said to be "irreversible"—structural changes on cooling will *not revert* on reheating until temperatures are reached many hundreds of degrees higher than the cooling change points. Hence, the alloy steels which retain mechanical properties at high temperatures become possible—high speed tools, engine valves and seatings, etc.

Besides pulling the A_c and A_r points down, pushing them up, and separating them, the alloying elements may also have the effect of pushing the points sideways to the left by inducing the formation of the eutectoid containing less carbon than the 0·9 per cent of the straight carbon steels. Chromium and tungsten either separately or together have the power of producing a eutectoid steel with as little as 0·2 per cent carbon; thus, even in small amounts, they increase the amount of pearlite for any given percentage of carbon.

Chromium and tungsten steels containing very little carbon may be hyper-eutectoid, and the free alloy cementite (composite carbide) is not a dangerously embrittling constituent like the plain cementite; tiny hard equiaxed granules or visible particles of alloy cementite may lurk undissolved, or "unsolutioned," in any of the transition products or in austenite without doing the mechanical properties any harm, often even improving the cutting qualities of the tool steels in which their presence is welcomed.

Composite Carbide Molecules as Lattice Pegs

Alloy martensites are not necessarily stronger and harder because the alloy carbide is a more effective peg in the lattice of the solid solution; method (iii) has done its good work on the solid solution, and the alloy carbide molecules are pegs at least as effective as the plain cementite pegs, perhaps better; and moreover, the steel diagram and hysteresis have been altered in various ways to permit us to ring many more valuable changes on heating and cooling.

Sluggishness

These alterations in the diagram by hysteresis may be explained on a theory of sluggishness which seems to meet the case. The alloying elements, depending upon their amount in the iron lattice, seem to produce a relative immobility of the lattice atoms, hindering carbide molecule formation in martensite, and checking the subsequent segregation, crystallization, and flocculation effects upon which the corresponding formation of troostite, sorbite, and pearlite depend.

This sluggishness with which changes take place in the alloy steels not only makes the changes late on cooling and late on heating—thus causing considerable hysteresis—but slows down the tempo of all changes which are really too fast to control nicely in the case of the straight carbon steels. It stands to reason that the formation of a metallic compound (like the carbides) from austenite calls for a certain amount of rearrangement: first a rearrangement to produce the alpha lattice from the gamma lattice, and next a rearrangement of certain iron atoms which unite with carbon and other atoms to form molecules of composite carbides. Now, if the alloying elements have reduced the mobility of the atoms in the lattices, all this rearrangement will be delayed and will take longer to perform after it starts; with all this delay which can be deliberately induced by using more and more of the alloying elements, the labile shower of molecular carbides (martensite formation) can even be checked until a considerable degree of cold may be needed to bring it about—when we have an austenitic steel, a steel of austenitic structure on cooling in air. Intermediate amounts of the alloying elements would provide steels in which martensite or troostite were the "normal" air-cooled structures, but from which austenite or martensite (respectively) could be obtained by rather faster cooling, or austenite (in both cases) on quenching. Small quantities of alloying elements might give sorbitic or pearlitic structures as normal structures, but rather faster cooling would produce troostite or sorbite, whilst quenching would produce austenite, martensite, or troostite, depending on the amount of the alloying element and the rapidity of the quench.

Cementitic Steels

Unlike plain carbide of iron, free or pro-eutectoid *alloy cementites* are not the brittle carbides crystallizing in grain boundaries, or occurring in tufts of crystals which constitute a danger to an alloy; they crystallize in tough granules which are exceptionally hard, and do no harm to the properties of any of the steel structures which may contain them through improper assimilation, or incomplete "solutioning." In consequence, it is of relative unimportance in the case of the alloy steels if the percentage of carbides is too great for the production of perfectly homogeneous austenite, for, if all the carbides are not dissolved in the gamma iron, those remaining are harmless either in the austenite or in any of the transition products.

Such carbides, in excess of the amount required to give a fully saturated solution of carbon in the austenite at any particular temperature, give rise to what are known as "cementitic" steels, and to the cementitic structure in steels. Beyond endowing the tool steels with improved cutting qualities, it could not be said that such occurrences of carbide effect the steels much when in relatively small amounts.

It can therefore be claimed for the alloying elements that they remove the danger of excess cementite: this may be one of the reasons why it has been found possible to produce very effective and satisfactory steel castings direct from the foundry cupola from high grades of blast furnace pig and the necessary alloying elements, eliminating the steel furnace entirely, and relying on an elaborate heat treatment of the casting to produce the desired steel structures and properties.

High Speed Tool Steels

These contain large amounts of alloying elements like chromium and tungsten so that but little carbon gives rise to a hyper-eutectoid steel and as much as 0·6 per cent provides a large amount of free alloy cementite (chiefly tungsten carbide). Up to 1260° C. all this cementite is not split up and put into solution, so that the austenite is not homogeneous but contains cementite granules. In subsequent heat treatment these granules remain unaltered whatever the product obtained.

The heat treatment (for best properties at high temperatures on the cutting edge) is to cool in an air blast to trap nearly all of the cementitic austenite at 1260° C., and by this means only a little martensite makes its appearance. Tempering at 600° C. then has the effect of changing much of this austenite to martensite which is now embedded in the sorbite formed by the tempering of the original martensite. It requires the high tempering temperature of 600° C. to effect this change; and, once austeno-martensite is formed, a still higher temperature than 600° C. will be needed to alter these structures and their properties substantially. Hence such a tool steel now consists of alloy carbides as granules embedded in austeno-martensite, which in turn is embedded in sorbite.

Stellites

A class of alloy known as "Stellite," developed for tool tips, valves and valve seatings where great strength and hardness are required at high temperatures, contains little or no iron and carbon, but consists substantially of cobalt, chromium, and tungsten, and may also contain manganese, molybdenum, and silicon.

These are cast alloys, and are weldable but not heat-treatable.

Cemented Carbides

The hardest known products of metallurgy are the so-called "cemented carbides" which bid fair to revolutionize the toolmaking industry. A typical substance of this class is prepared by pressing an intimate mixture of tungsten carbide and cobalt metal fine powders into a mould, and sintering them together under heat and pressure, or merely by heating after the pressure has consolidated the powder.

Alternatives to tungsten carbide are titanium carbide or tantalum carbide, and the alternative to cobalt is nickel. Cemented carbide products differ in their content of the above carbides and metals. Frequently, the sintering is in two stages, after the first of which the product is still not too hard for shaping with tools. A subsequent higher temperature application of 1500° C. or so completes the considerable shrinkage and densification, whilst hardness reaches extraordinary heights (corresponding to about 1400 V.P.N. or true Brinell) well above that of the Stellites and largely retained at very high temperatures.

Such materials are used as tool tips for drawing dies and wherever resistance to corrosion and abrasion is to be very high.

CHAPTER XV

ALLOY STEELS

The Effect of Individual Alloying Elements

To consider first the elements which are usually present in small quantities in straight carbon steels because they commonly occur in pig iron—

Silicon

Small quantities have strong effect, going into solution in the iron, and not forming carbides. It tends to inhibit carbide of iron formation and favours the formation of graphite instead.

Silicon additions raise the elastic limit, strength and hardness, but seriously reduce ductility, toughness and notch toughness. Therefore, as a solid solute in iron, it is not altogether a success from the standpoint of mechanical properties. It is a cheap strengthener, however.

In common use are—

Silico-manganese Steel for leaf springs, 0·5 per cent C, 1·5 per cent Si, and 0·8 per cent Mn.

Silicon Steel for transformer cores and the poles of dynamos and motors—for high permeability and resistance (electrical). Lowest possible carbon percentage, 4–5 per cent Si.

(This has very poor mechanical properties.)

Silchrome for automobile valves—

0·4 per cent C, 4 per cent Si, 8 per cent Cr.

(Tough only at higher temperatures.)

Silicon raises both A_c and A_r points.

Manganese

Manganese partially dissolves in the iron and partially forms Mn_3C which is found in the alloy steel cementite together with the Fe_3C.

Manganese lowers the A_r3 and A_r1 points and has the same effect as an increase in cooling rate, so that even small quantities give a normal structure (air-cooled) of *sorbitic* pearlite.

Above 1·8 per cent Mn the normal steels tend to be martensitic or "air-hardened" on account of this effect on cooling rate, and to suffer considerably in ductility.

Between 5 per cent and 12 per cent Mn the steels are largely martensitic (even after slow cooling) and are useless commercially.

Hadfield's Manganese Steel, 12 to 14 per cent Mn and 1 per cent carbon, is entirely of austenite upon quenching from 1000° C., in which soft condition it is used; but, when quenched, it is metastable and immediately takes on a "strain-hardened" surface under the effects of abrasion or pressure, etc.

The mechanism of this particular strain-hardening is worthy of note; for cold work does not, in this case, produce "amorphous material," as would be the case in cold working an equilibrium product such as a pure metal or solid solution in equilibrium at ordinary temperatures. Hadfield's Manganese Steel is a *quenched* product which cold work causes to revert to its equilibrium product, martensite.

This is only one of the well-known instances of cold work producing an equilibrium from a metastable product.

Nickel

Though forming no carbide, this element somewhat resembles manganese in its effects, but the A_r points are lowered more rapidly than the A_c points with increases of nickel until, at about 8 per cent, the A_r points take a plunge causing the "hysteresis" or lag between the A_c and the A_r points to be considerable.

A steel containing 12 per cent nickel begins to transform (A_r3 point) below 300° C. on cooling; but, on reheating, the reverse change does not occur until about 650° C. Such steels are called "irreversible steels."

Additions of nickel have the effect of increases in the rate of cooling of a carbon steel: up to about 8 per cent nickel, air-cooled steels are sorbitic or troostitic: from 8 to 10 per cent nickel, the structure is martensitic: above 24 per cent nickel, the A_r3 point is depressed below room temperatures and the air cooled structure is austenitic—a proper austenitic steel.

Nickel produces a fine grain and therefore a high notch toughness; it also produces a fine pearlite, and gives extra strength not only on that account but because the nickel bearing ferrite (solid solution) is itself stronger; there is little falling off in ductility until about 8 per cent nickel is present; so this is a very valuable solid solution indeed.

Carbon intensifies the action of nickel in a manner shown well in the Guillet Diagram.

Case-hardening steels containing nickel may contain carbon 0·1 per cent and nickel 2 to 5 per cent, but the principal pearlitic (and sorbitic) nickel steels are those containing 0·25 per cent to 0·4 per cent carbon and 3 to 3·75 per cent nickel. These are used for crankshafts, axles, and connecting rods.

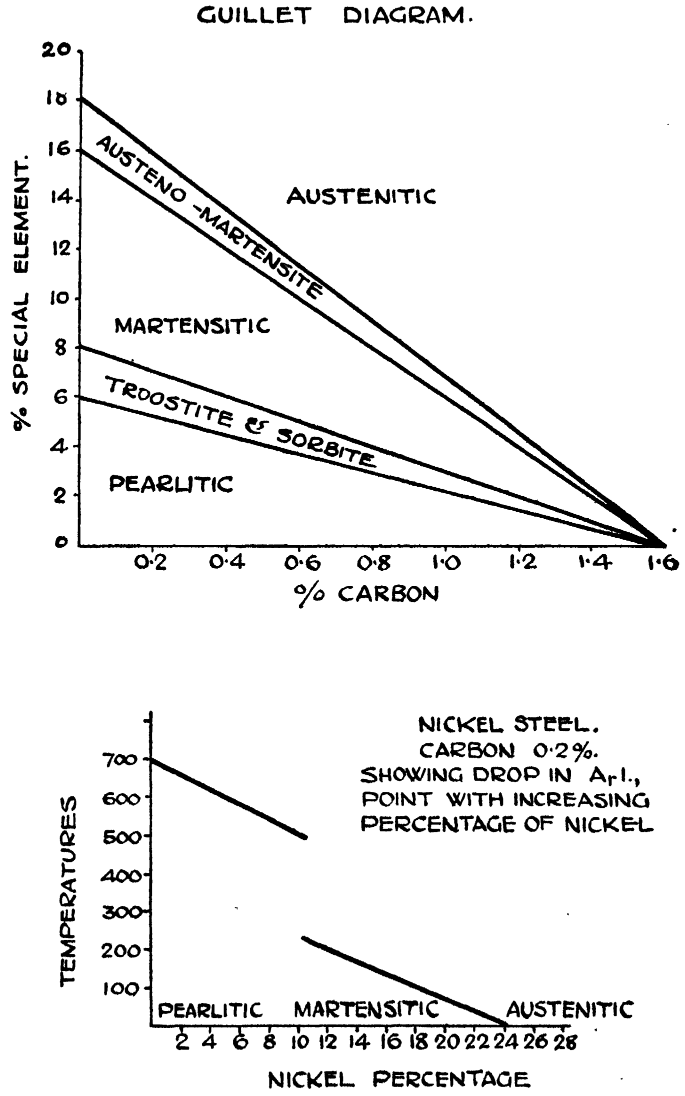


Fig. 23

The best properties of these pearlitic steels are brought about by quenching and tempering. The A_c3 point is lower, hence a lower hardening temperature than for straight carbon steels is permissible: moreover the range of hardening temperatures without undue grain growth is larger.

Martensitic nickel steels are not used.

Austenitic nickel steels cannot compete with austenitic manganese steels on account of high cost.

Invar

Invar is 0·2 per cent C, 0·5 per cent Mn, and 36 per cent Ni and its coefficient of expansion by heat is practically nil between 0° C. and 350° C.

Chromium

Chromium forms solid solution and carbide as well (Cr_3C_2); the latter combines with Fe_3C, forming very hard cementite which is "stable," i.e. only slowly dissolved by austenite and slowly precipitated from it.

This element raises the A_c points and also the A_r points with slow cooling; but A_r points are lowered with fast cooling on account of the above-mentioned stability. As it promotes grain growth, another element such as nickel is commonly added to inhibit this growth.

Chromium induces hardening not only on account of its hard solid solution, but because of its troostite and martensite formation on cooling (unless the cooling is very slow). Moreover, the carbon percentage of the pearlite is lowered so that for any carbon percentage there is a greater amount of pearlite in a chromium steel than in a straight carbon steel.

Above about 10 per cent chromium the steels are possessed of inert corrosion-resisting films; high chromium contents provide steels with considerable immunity from oxidation and from loss of properties at the higher temperatures.

Easier to machine than nickel steels of similar strength, these alloy steels suffer from brittleness, on tempering in a low range of temperatures, known as "temper brittleness," which is low notch toughness in spite of good ductility.

Dies, ball bearings, plates for safes, rolls, files, and tools, call for the extreme hardness which chromium provides. Certain permanent magnets are made of chromium steels.

Nickel and Chromium together in steel combine all the good properties, the nickel correcting the grain growth due to chromium and the chromium inhibiting graphitization and keeping the critical points up.

The depth of hardening is increased when these two alloying elements occur together.

Molybdenum

Molybdenum forms both solid solution with iron and carbide with the carbon; it reduces the critical range on cooling, and also lessens the softening effect of tempering.

This element alone has been used in boiler steels to give increase in strength at high temperatures (400° C.), but it is chiefly used with nickel and chromium to ensure the absence of temper brittleness, and to reduce oxidation.

From 0·3 to 1 per cent or more of the element may be used: it is three times as powerful as tungsten in producing very similar effects, but it is very much more expensive than tungsten.

Vanadium

Vanadium is often added to ensure the freedom of the steel from oxides; thus it scavenges and cleans the steel. It forms a carbide, improving the properties of heat-treated steels.

0·15 per cent V appears in the chromium-vanadium steels used for locomotive forgings, automobile axles, coil springs and torsion bars.

Tungsten

Tungsten forms a solid solution with the iron, and a hard carbide (WC) with the carbon. It raises the critical points, and the carbides dissolve slowly over a range of temperature: when completely dissolved, the transformation to pearlite is sluggish and air hardening results, i.e. air cooling lowers A_r points as in the case of chromium.

The hardened steels therefore resist tempering to a higher temperature than ordinary steels, and use is made of this for (hot working) tool and die steels. It refines grain and helps to avoid decarburization during hot working.

It is also used in magnet steel and in corrosion and heat-resisting steels.

Copper

Copper dissolves in iron to a limited extent; about 3·5 per cent is soluble in iron at normalizing temperatures but only 0·35 in ferrite at ordinary temperature. Resistance to atmospheric corrosion is improved by copper; and "temper hardening" of copper steels can be carried out. In temper hardening hardness and strength are increased by tempering after quenching. The quenching gives a comparatively weak and soft product which is unstable: the tempering induces the equilibrium product—or a step towards its formation—which is stronger and harder. In this case copper molecules (groups of copper atoms) form, and these peg the lattice more than the copper atoms.

CHAPTER XVI

THE NON-FERROUS METALS AND ALLOYS

THE physical metallurgy of the iron, or "ferrous," alloys is so extensive that it is dealt with almost as a separate subject. It should, however, be realized that there is no real significance in the classification of metals and alloys into ferrous and non-ferrous.

Having seen what steels are, we are now to find that precisely analogous and similar effects of every kind known in steels may be induced also in alloys containing no iron and no carbon, but only copper, tin, aluminium, magnesium, etc.

Methods (iii) and (iv) (of Chapter VII) work as surely (though not, perhaps, as effectively) if the copper lattice, for example, is pegged with aluminium units to form an "austenite" at high temperatures from which at lower temperatures a "pearlite" or eutectoid forms, containing an alpha copper phase and a *compound* of copper and aluminium.

But we should not be widely understood if we applied the names "austenite" and "pearlite" to the identical structures of the non-ferrous alloys; so we must avoid such attractive designations in deference to the conventions governing the metallurgists of to-day.

All ferrous alloys are not steels, for some contain little or no carbon, and others are not treated to obtain the well-known structures of the steels. On the other hand, some non-ferrous alloys (like the above) are more "steels" in a real sense than many ferrous alloys.*

Structural and Non-Structural Alloys

A certain classification of all alloys should assist in securing a correct outlook on alloys *vis-à-vis* their mechanical properties—

Class i. Those "structural alloys" designed primarily for the display of certain mechanical properties, like the steels, bronzes, and the like.

Class ii. Those designed *primarily* for the display of certain other properties, like bearing metals and non-corrodible metals, but which are nevertheless required to have the best mechanical properties in the circumstances.

---

* Copper alloyed with small quantities of beryllium and cobalt has been found to produce some compound on heat treatment which raises the U.T.S. to 80 tons per sq. in. or more with a percentage elongation of about four.

Class iii. Those designed for the display of some electrical or other such property hardly calling for mechanical properties worth the mention.

Keeping the above in mind, the engineer should not be so puzzled to find that bearing alloys—containing metals and constituents which are woefully weak and inductile by the standards of the structural alloys—may be described as "strong" and "plastic" by the standard of non-structural alloys. Nevertheless, some of the bearing metals nowadays are bronzes or lead-bronzes of strength and ductility comparable with those of the structural alloys.

The above classification is well worth making just before discussion of the Bearing Metals, more particularly since there remain to be considered the non-ferrous metals and alloys so freely used in Classes ii and iii as well as in Class i.

BEARING METALS

General

Particular properties are required of a group of alloys for machine bearings; whereas strength and ductility of a certain order are certainly called for in such alloys, it should be realized that this "order" is not the same as for structural alloys. The strongest bearing metal would, as a rule, be unsuitable in structures; and, probably, the plasticity called for in a bearing metal is more of the nature of a "creep" than a true ductility.

A bearing metal should, if possible, have—

Low coefficient of friction.

A surface of a heterogeneous nature—harder parts embedded in a softer matrix which, in wearing down, gives rise to harder parts standing in relief. In this way the oil film is substantially unbroken under conditions of use.

A plasticity sufficient to enable the bearing to adjust itself to any malalignment.

The hard parts must not be hard enough to scratch the shaft surface, and the matrix must be firm enough (when backed by structural alloys) and yet soft enough to wear preferentially, and to permit accidentally introduced particles of grit to sink down into it out of harm's way.

N.B. The lowest coefficients of friction are now believed to exist between metallic surfaces which consist of "amorphous" *metal* (or amorphous oxide). In a carefully run-in engine cylinder, the steel liner is found to have a surface of "amorphous" metal for a considerable depth. Certain hard constituents are found to be more suitable than others owing, possibly, to their capacity for holding, at their crests in the bearing surface, metal in the near-amorphous form, or else an oxide in the truly amorphous state.

The well-known low coefficient of friction of cast-iron is due, however, to self-lubrication of a very effective kind by its own graphite: but the surprisingly high coefficient of friction between an aluminium piston and a steel liner seems to be undoubtedly due to the *crystalline* (not amorphous) cutting effect of the aluminium oxide film.

Depending largely upon the loads and speeds, heat to be dissipated, and temperature of running, bearing metals may be of a variety of structures and properties as follows—

White Metal (lead–tin–antimony), including Magnolia Metal (Pb–Sb).
Babbitt Metal (copper–tin–antimony).
Lead or Plastic Bronzes (copper–tin–lead).
Bronzes (copper–tin).

White Metal (Lead–Tin–Antimony)

The binary alloy of lead and tin should be studied: lead-rich alloys at room temperatures consist of finely divided lead and tin—just soft lead grains, containing a fine dispersion of harder tin particles, set in a fine matrix of lead and tin eutectic. (See diagram.)

The binary alloy of lead and antimony at room temperatures consists of grains of lead (or else of antimony) and a moderately hard and brittle eutectic of excessively finely mixed lead and antimony. (See diagram.)

On the other hand *tin and antimony* form a hard (brittle) compound together, and cuboid crystals of these readily form when over about 7 per cent antimony is present. (See diagram.)

For line shafting, an inexpensive alloy would consist of 8 per cent tin, 20 per cent antimony, and 72 per cent lead.

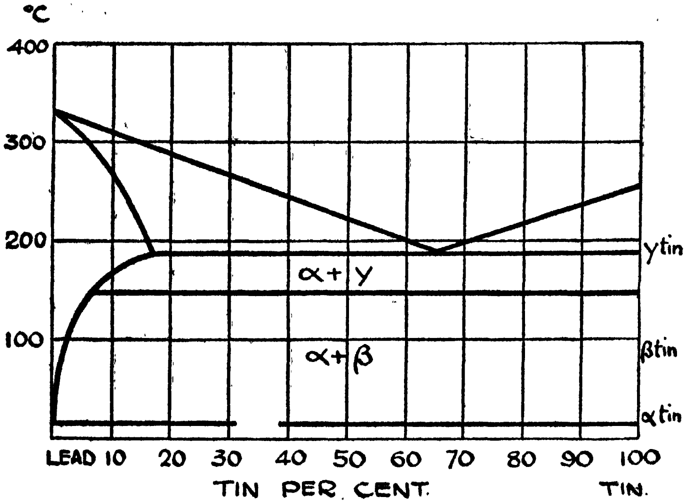
These three metals are present in some White Metals in the proportions necessary to provide a quantity of the cuboid compound crystals as the hard bearing constituent; an adequate matrix consisting of the **lead-antimony** eutectic and lead is present.

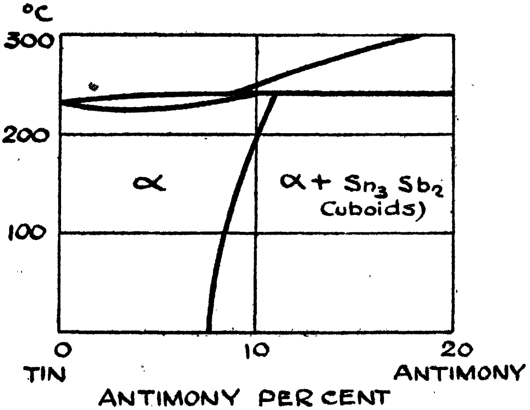
For very light loads, or in the case of long bearings with medium loads, antimony metal itself is used as the hard bearing constituent in a eutectic of lead and antimony—as in Magnolia Metal.

Babbitt Metal

For heavier loads some copper may be introduced, and much tin instead of the lead to produce a tin-rich bearing of Babbitt Metal.

LEAD-TIN DIAGRAM





(PART OF THE TIN ANTIMONY DIAGRAM)

Fig. 24

87 per cent tin, 10 per cent antimony, 3 per cent copper is an example in which all the lead has been replaced by tin.

In this case the hard constituents are the cuboid SbSn compound as before together with star-shaped crystals of the copper-tin compound $Cu_6 Sn_5$. The softer wearing matrix is now a eutectic of tin-antimony solid solution + $Cu_6 Sn_5$, which latter forms a fine network keeping the SbSn cuboids properly distributed during solidification.

Plastic Bronze Bearings

Ten to 30 per cent of lead can be emulsified in alpha tin-bronze, particularly if about 1 per cent of nickel is present as well.

Bronze Bearings

Twelve to 18 per cent tin bronzes contain the eutectoid which is the hard constituent of such strong bearings. (See Tin Bronze.)

Bearings by " Powder Metallurgy "

Bearings are also made by moulding mixtures of powdered copper, tin, and graphite under pressure around 20 tons per sq. in. and afterwards sintering in a reducing atmosphere.

CHAPTER XVII

THE COPPER-BASE ALLOYS

THE "bearing metals" just considered are non-ferrous, but they belong to a group used for one particular purpose.

Non-ferrous alloys for structural and general purposes must possess the highest possible strength, hardness, etc., compatible with the necessary toughness, notch toughness, corrosion resistance, cold or hot workability, electrical and other properties which may be desired.

Copper-base Alloys—the Brasses

Next to the large demand for iron ranks that for copper. This is to some extent owing to the ease with which it may be drawn into high conductivity wire for electrical apparatus of all kinds, but its use in the brasses and bronzes is of great importance.

Brasses—alpha and alpha-beta (α and $\alpha\beta$)

The solid solutions of zinc in copper are classical examples of the desirable effect of alloying metals to improve mechanical properties.

Zinc is soluble in copper in the solid state to the extent of nearly 50 per cent giving alloys remarkable for mechanical properties, ease of working, colour, and resistance to atmospheric and marine corrosion.

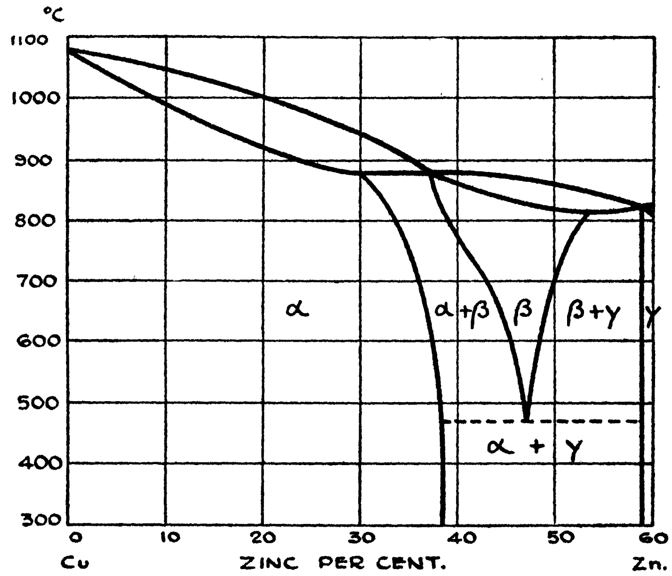
As the zinc content of the brasses is raised from 0 per cent to 30 per cent, brasses of ever increasing strength and ductility are produced; the products, cheaper than copper alone, are not only superior in mechanical properties, but are more easily worked either hot or cold by drawing, rolling, forging or extruding; they resist corrosion better, have lower melting points and are even lighter.

Beyond 30 per cent zinc the alloys still consist of the same α phase until 39 per cent Zn is reached (for an alloy in equilibrium at room temperatures), but the ductility now falls off to some extent whilst the strength, hardness, etc., continue to increase more sharply.

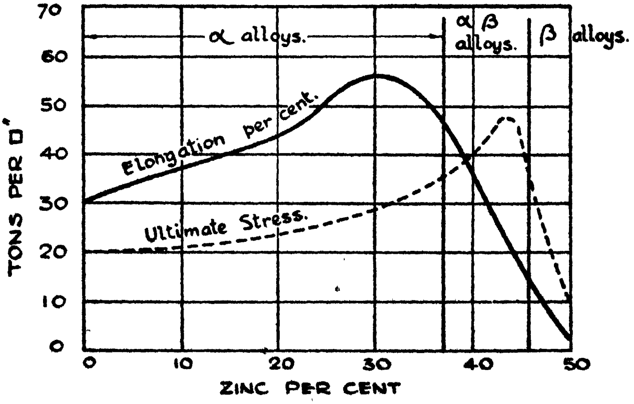
For cold working with exceptional ease, then, the limit of zinc is approximately 30 per cent; but the brasses up to 39 per cent can be cold worked with but little difficulty, and very easily indeed by hot work.

Further increases of zinc beyond 39 per cent causes the introduction of the β phase, so that between 39 per cent and 45 per cent

THE BRASSES DIAGRAM



(PART OF THE COPPER-ZINC DIAGRAM)



MECHANICAL PROPERTIES.

FIG. 25

zinc $\alpha\beta$ brasses occur in which the strength reaches its maximum of 32 tons per sq. in. (U.T.S.), though from 30 per cent Zn the ductility has been steadily falling. These $\alpha\beta$ brasses are hot worked with great ease.

Increase in ductility on forming solid solutions is not common though increases in strength, hardness, etc., are always to be expected owing to the pegging of the parent lattice by the units of the alloying metal (the solute).

Another uncommon feature of these alloys is the *increase* in solid solubility with decrease in temperature, as can be seen from the diagram.

Brasses of 50 per cent zinc contain a γ phase which renders the alloys extremely hard and brittle. Such an alloy would only be used in brazing for its low melting point and since some zinc will volatilize from it and some diffuse into the metals being united.

Annealed alpha brasses will withstand a remarkable degree of deformation by cold work (cartridge brass). There is no appreciable advantage in hot working an α brass; except that in breaking down large ingots into strip, repeated annealings become unnecessary if hot work is employed.

It is advantageous, however, to hot work a brass containing the β phase because there is an important change in this phase at 470° C. According to some reliable authorities, this change is *strictly* in line with that taking place when so many other alloys develop a compound from a solid solution on cooling; though the latest research has indicated a somewhat more complicated change not so suitable for the attention of engineers.

The impact resistance of α brass between 350° and 650° C. is certainly very poor though its hardness persists right up to 800° C. and its creep resistance is better than that of a β brass.

Dendritic (or cored) structure will appear in cast brass, but annealing will correct this.

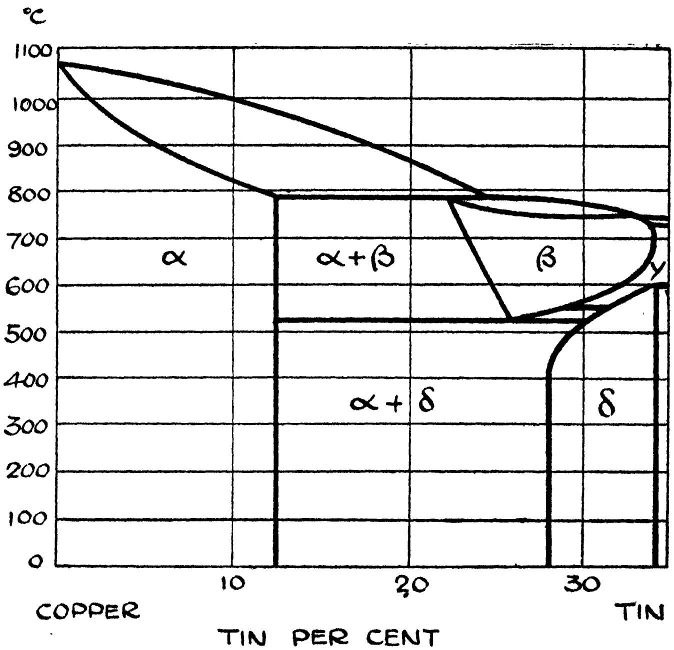
" Cast " and " Wrought " Alloys

All alloys are cast in the first instance, but if any type is put into use just cast or chill cast (with or without heat treatment) it is frequently described as a "cast" or a "casting" alloy; if, on the other hand, an alloy is commonly used only after extruding, forging, stamping, drawing or rolling, it is known as a "wrought" alloy.

Many alloy compositions are suitable both for cast or wrought finishes; but frequently a slight difference in composition will distinguish the alloy to be used as cast from that to be used after hot or cold working.

The brasses, bronzes, and nickel bronzes, are both casting and wrought alloys; the old Y alloy and modifications were aluminium-base casting alloys only, being unsuitable for working hot or cold; Duralumin is somewhat weak as cast, and is always used after extrusion and/or other hot and cold work. It is, therefore, always a wrought alloy.

TIN-BRONZE DIAGRAM.



(PART OF THE COPPER-TIN DIAGRAM.)

FIG. 26

Copper Base Alloys—The Bronzes

Tin is soluble in copper in the solid state up to 13 per cent in equilibrium alloys; but, owing to the wide temperature range between liquidus and solidus, lack of equilibrium conditions is common at rate of cooling usually accounted slow, so "coring" or "dendritic structure" is common in the bronzes. (Chapter VI.)

Therefore, if the plain structure of a solid solution is required under practical rates of cooling, the tin should not exceed about 7 per cent.

Beyond 7 per cent tin in copper the usual structure is a cored solid solution as the α primary, or as the β solution if the temperature is above 520° C.—which latter changes to an $\alpha + \delta$ eutectoid below 520° C.

Particular attention is drawn to the fact that this β to $\alpha + \delta$ change is similar in all respects to the change from austenite to pearlite (γ to α + Cem) in steels. The β constituent of bronzes may be regarded as a kind of austenite which, in its transition to $\alpha + \delta$ below 520° C., does not happen to display a martensitic-looking structure. On quenching, though, it certainly does exhibit transitional strength and hardness (etc.), which further tempering will reduce and ultimately remove altogether.

The tin bronzes then develop the same constituents as steel in the same way with the same kind of heat treatment, though they do not develop the visible structures of martensite, etc.

Aluminium Bronzes—10 per cent alloy for Castings or Hot-working

On the other hand the aluminium bronze (copper-base) alloys for casting or hot-working develop not only the same constituents as a steel in the same way by heat treatment, but also (it so happens) the same fine Widmanstätten structure* which is the martensitic structure due to the first formation of a high dispersion of δ compound on the octahedral planes of the α constituent. Here again, it is to be presumed that the highest dispersions of the δ constituent, i.e. the molecules of the compound, though sub-microscopic and invisible, nevertheless play the most important part in the strengthening and hardening by their powerful pegging of the lattice; dispersions of a lower order have less effect in modifying the parent lattice properties, until ultimately a full anneal will freely flocculate the compound so that it is easily visible, or coarse, when it no longer has any effect in strengthening the parent lattice. This alloy is quenched at about 900° C. and tempered between 200° and 550° C. For full details see the Heat Treatment Table.

The Wrought Alloy

The wrought alloy of aluminium bronze contains 5 to 7 per cent aluminium: it is easily worked, hot or cold. As cast this alloy is cored, but has only one constituent and the common (twinned) solid grains are produced after working and annealing.

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* As in the case of steels, the Widmanstätten structure of the martensite may equally be due to the β to α change actually prior to any compound formation.

Lead in Tin Bronzes

Insoluble globules of lead are formed. Up to 2 per cent of lead is added to bronzes to improve machining.

Larger quantities reduce strength and ductility, and corrosion resistance.

From 8 to 30 per cent lead may be added to bronzes for bearings; these are called "plastic bronzes": 1 per cent nickel assists in giving a uniform distribution of the lead globules in the mass.

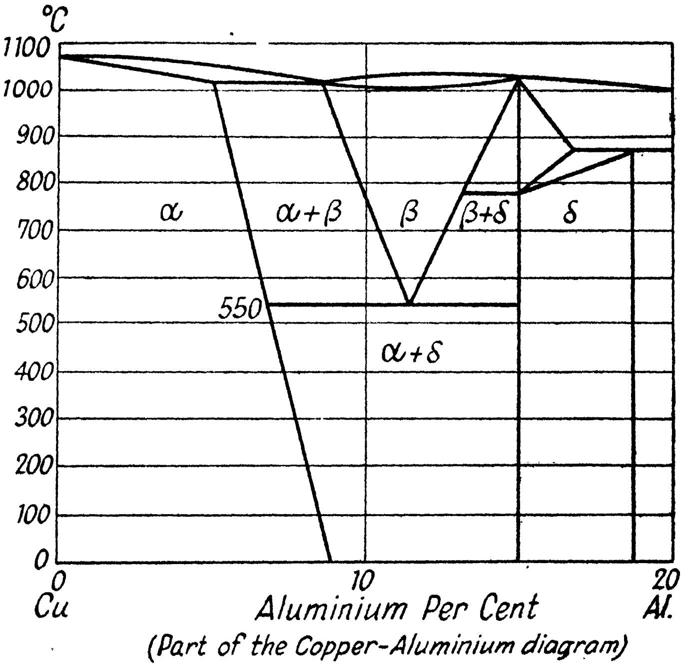


FIG. 27. THE ALUMINIUM-BRONZE DIAGRAM

Phosphor-bronze

Phosphorus is a powerful deoxidizer and additions of it (as phosphor-copper, 15 per cent P) are made to the copper before the additions of the tin. In this way the copper oxide, which is prone to occur dissolved in the copper, is removed before tin has a chance of combining with this oxygen to form the very troublesome tin oxides.

Only traces of phosphorus remain in the alloys called "phosphor-bronzes" if the phosphorus is used as a deoxidizer only; but if over

0·1 per cent phosphorus remains, it has a definite effect on structure and properties in the resulting alloy, imparting improved strength and resistance to sea-water corrosion to bronzes in the form of rods, sheet and wire; it provides an excellent bearing surface for heavy loads, and also gear wheels and slide valves as bronze castings.

Copper-Nickel Alloys

These are entirely different to the brasses and bronzes, for these two metals form together a complete and uninterrupted series of solid solutions between 100 per cent copper and 100 per cent Ni.

80-20 cupro-nickel is noted for its extreme malleability in the cold: corrosion troubles in condenser tubes can largely be eliminated by the use of this alloy which is also free from "season-cracking" (corrosion of amorphous material of cold work, and breaking up under the residual internal stresses left by cold working).

Monel Metal contains 68 per cent Ni and 2 per cent Fe: it has a U.T.S. of 37 tons per sq. in. and a ductility of 40 per cent. Resistant to superheated steam, and very retentive of strength at high temperatures, it may be used for turbine blades, valve parts, pump rod liners and impellers. **K Monel** contains aluminium as well, and is heat-treatable with very excellent results owing to some compound formation.

Nickel Silvers, or German Silvers, are copper-nickel-zinc alloys, which are special brasses of excellent properties. Automobile radiator shells, fittings of all kinds, spoons and forks (for electro-plating with silver or chromium), wire and tape for electrical resistances.

CHAPTER XVIII

ALUMINIUM-BASE ALLOYS

PURE aluminium is a rarity: commercial aluminium is very weak in the annealed condition and only moderately strong (up to 10 tons per sq. in.) in the fully rolled state.

Aluminium is never used alone in castings, but always alloyed.

Casting Alloys of Aluminium

Aluminium is an expensive metal; wherever weight and expense are to be avoided in the manufacture of unstressed parts, alloys of aluminium base with 12 to 14 per cent of zinc and 3 per cent or so of copper are cast, or chill-cast, or hot and cold rolled.

As cast or chill cast there is a serious falling off in the ductility of these alloys compared with that of aluminium, but the strength, hardness, etc., may be more than double those of aluminium if hot or cold work treatment is applied.

Y Alloy—for Casting and Heat Treatment

Four per cent Cu, 2 per cent Ni, 1½ per cent Mg. (with Si and Fe as accidental impurities) has great toughness and strength at high temperatures, and it is particularly good for pistons. The Blackburn Cyrrus cylinder head is made of Y alloy.

It is not an easy composition to cast.

Heat treatment consists of "solutioning" at 500° C. and quenching in boiling water followed by a period of ageing, or else quenching in oil and ageing artificially at 160° C.

(Aluminium-base) Casting Alloys

The strength and toughness induced by the specified heat-treatment are due to the following submicroscopic structure development.

At a temperature of 500° C. the copper (4 per cent) and the magnesium (1½ per cent) are in solid solution in the aluminium as well as the nickel (2 per cent) and the Fe and Si.

Equilibrium cooling rates from this temperature would result in the formation of molecules of compounds of $Cu\,Al_2$, $Mg_2\,Si$, $Ni\,Al_3$ within the aluminium lattice and their subsequent gradual *flocculation* to masses visible under the microscope. Excessively tardy cooling rates would yield a fully annealed structure of very coarse *segregations* of these compounds and result in an alloy almost indistinguishable from aluminium in mechanical properties.

But the specified heat treatment is a quench at 500° C., whereby an unstable product is trapped; this product is a "supersaturated solid solution" which "ages" in time at room temperature, or which can be aged more rapidly at some elevated temperature (such as 160° C.) for a very short time of "precipitation."

Such ageing is simply the development, from the plain solid solution, of a *molecular dispersion* of these compounds throughout the aluminium lattice. The ordinary proportion of the total available compounds actually existing as a *molecular* dispersion in age-hardened Y alloy (or Duralumin) is at present unknown; but it would seem that the maximum attainable proportion is not too much —that is to say, does not produce strength and hardness of such a high degree as to jeopardise ductility (as in martensitic steel).

Though *natural* age-hardening occurs at ordinary temperatures over a long period of time, and *artificial* age-hardening occurs at fairly low temperatures (such as 160° C.), nevertheless the structures obtained by either method of ageing are permanent at relatively high temperatures such as those reached at the top of a Y alloy piston—possibly 300° C. or more.

In fact, higher temperatures are required to *flocculate* and segregate the compounds which have been formed (or precipitated) by the age-hardening treatment; hence higher temperatures than 300° C. would damage the properties which accompany the highly dispersed structure, and higher temperatures would be needed, likewise, for annealing to soften. The alloy quenched at 500° C. is in its softest state, but cold working it in this condition might soon age-harden it. The alloy should be annealed before working it cold.

Quenched Y alloy (or Duralumin) then changes to something like a troostitic analogue (without the intervention of the hard brittle martensitic analogue) *naturally* at room temperature over a certain long period of time. This time can be cut short if artificial ageing is carried out. Having settled down to this troostitic equilibrium at room temperature (or somewhat higher temperature) further heating will have no effect on structure or properties until high temperatures such as 300° C. are reached; then flocculation or segregation of the compounds begins: the alloy is softened, and ultimately put into the annealed state.

To revive the good properties of the over-heated alloy, all that is necessary is to repeat the treatment: quench at 500° C. and allow to age-harden naturally, or artificially age-harden by keeping at 160° C. for a short time.

Aluminium-base Casting Alloys

Silicon Alloy—Alpax, Wilmil, etc.
(11 to 13 per cent Silicon)

Silicon is expensive, but it is light and extremely effective in producing a really good casting alloy with aluminium. Such an alloy can easily be cast in $\frac{1}{8}$ in. sections but should not be used for stressed parts. (U.T.S. 8 or 9 tons per sq. in., elongation 0·5 per cent.)

There is still some mystery surrounding the questions of ordinary and "modified" silicon alloy; for, by mixing or dissolving a very small quantity of sodium (0·05 per cent) with the alloy, improved strength and toughness result (up to 13 tons per sq. in. U.T.S. and 15 per cent elongation). The structure of these alloys is that of the eutectic of aluminium and silicon, but the sodium appears to increase the silicon content of the eutectic and to induce a fine grain-size giving rise to the "modified" structure.

Additions of 0·3 per cent Mn and 0·3 per cent Mg improve the proof stress which is otherwise poor.

High Duty Alloys—Alloys of the R.R. Series

In the laboratories of the Rolls Royce Company and elsewhere, improvements have been effected upon the original Y alloy to which iron, silicon and titanium have been added to produce the **Hiduminium Series,** for crankcases and supercharger cases, etc. (16–19 tons per sq. in. U.T.S., Proof Stress almost as high.)

Adaptations of these are also used as wrought alloys.

Aluminium-base Wrought Alloys

Duralumin, the first heat-treatable non-ferrous alloy to be discovered, is still widely used for general purposes as forgings, stampings, bars, sheets, tubes, and rivets.

An average composition is—

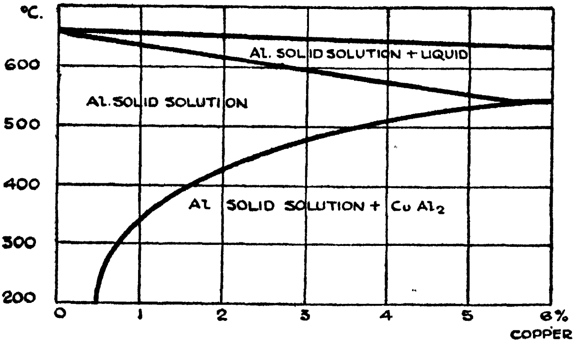
Cu, 4 per cent; Mg, 0·5 per cent; Si, 0·5 per cent; Fe, 0·5 per cent; Mn, 0·5 per cent.

The iron and silicon are natural impurities, but sometimes a little additional silicon is deliberately added.

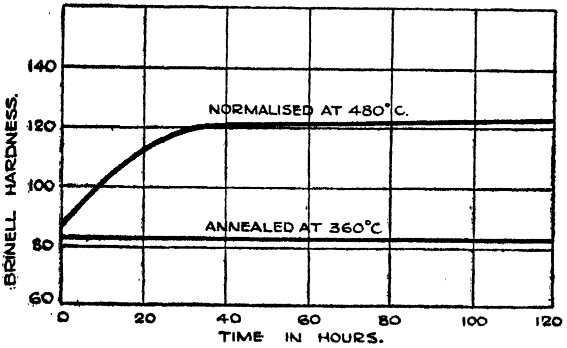
Somewhat brittle as cast, it is preferably hot worked at 450° C., after which it can be cold worked to some extent.

Its heat treatment resembles that of Y alloy, in which Mg_2Si and $CuAl_2$ compounds play prominent parts. Careful investigation has shown that when these two components occur together they limit each other's solubility in aluminium severely, so that there is no solubility of either at 250° C., but only above that temperature.

DURALUMIN AND Y ALLOY.



PART OF THE
ALUMINIUM-COPPER CONSTITUTION DIAGRAM.



THE AGE-HARDENING OF DURALUMIN.

FIG. 28

Solutioning occurs between 400° and 500° C. (which temperature should not be exceeded). Quenching fixes a soft ductile supersaturated solution of Cu, Mg, Si, etc., in aluminium from which, quite spontaneously in the course of time (ageing), a labile shower of the compounds Mg_2Si and $CuAl_2$ is formed.

Ageing at room temperature is rapid at first, but requires several days to reach its maximum value.

Artificial ageing can be carried out from 100° to 150° C.: this merely hastens the labile shower formation, and saves time.

Heat treating to temperatures between 300° and 380° C. has the effect of flocculating the compounds, rendering them visible under the microscope and softening the alloy: this is the annealing range, but annealing and change of properties do not take place readily below 350° C.

Worked, quenched, and aged, Duralumin has the U.T.S. of 27 tons per sq. in. and an elongation of 20 per cent.

Duralumin H contains no copper but relies on Mg_2Si hardening solely. There is but little tendency to age at room temperatures, and artificial ageing at 160° C. commonly follows water quenching from 520° C.

R.R. Alloys are modifications of Duralumin by increasing the Fe, and by additions of Ni and titanium; these increase the U.T.S. to 29 tons per sq. in. with elongation 15 per cent and Proof Stress 24 tons per sq. in.

CHAPTER XIX

MAGNESIUM-BASE ALLOYS, INCLUDING "ELEKTRON" ALLOYS

THE lightest commercial alloys available, being bulk for bulk two-thirds the weight of aluminium and one quarter the weight of steel. By strength/density ratio, Elektron is 11·1; heat-treated nickel-chromium steel, 7·0; mild steel, 3·6.

Usually contain 8½ per cent Al, 3½ per cent Zn, and ½ per cent Mn; or else 10 per cent Al and 1 per cent Zn only.

(Molten magnesium easily catches fire, but burning may be prevented by the use of a flux of magnesium chloride, calcium fluoride, and magnesia. The attack of molten magnesium upon the sand of moulds may be prevented by using 5 per cent of sulphur with the sand.)

As cast the U.T.S. is about 9 or 10 tons per sq. in., elongation 2 to 4 per cent. A solutioning anneal at 400° C. for several hours followed by an artificial ageing at 160° C. to 210° C. has some effect in improving these properties.

For the fabrication of fuel tanks and cowlings in aircraft work an alloy containing about 1 per cent Mn is used: it can be welded, and should be bent when hot (300° C.).

Elektron and similar alloys can be used either cast or wrought. An alloy—

8 per cent Al, 0·5 per cent Zn, and 0·2 per cent Mn

is particularly suitable for airscrew blades and crankcase forgings; for these purposes a cast billet is extruded, rolled and forged hot. All working of Elektron must be done hot between narrow temperature limits, 270°–330° C. Of this product the U.T.S. is about 20 tons per sq. in., and the elongation 10 to 15 per cent.

"High Duty" Alloys

Are those of the light alloys which have been improved, not by new principles of solutioning and precipitation, but by attention to details of grain-size, casting efficiency, and porosity.

Small quantities of titanium in the solid solutions of the light alloys keep the grain fine. Up to 1½ per cent of iron may be used for the same purpose. A little manganese is generally effective in improving "founding" properties, probably by removing oxygen and thus keeping the molten alloy limpid.

Nickel additions have been found to reduce porosity; and, as porosity is thought to be due to finest inclusions of oxide and nitride throughout the crystal boundaries, nickel also seems to be capable of deoxidation. A nickel compound (or compounds) occurs and plays a part similar to $Cu\ Al_2$ and $Mg_2\ Si$.

Metallurgists will, no doubt, arrive in time at more valuable and perhaps distinctive products by pursuing this method of trying small additions to improve the older products.

The alloys here mentioned are only a very few of those in more common use; thousands of different compositions are employed to meet specific requirements, yet little may come to light about these until some important development draws attention to the great amount of work being done in the laboratories and foundries.

CHAPTER XX

HEAT TREATMENT FOR OPTIMUM PROPERTIES

WHILST it must already be obvious that a general similarity exists in the heat treatment applicable to ferrous and non-ferrous alloys for the purpose of obtaining the best compromise between strength and ductility, only such a table as the one here shown can prove the identity in principle and in reality of a somewhat confusing mass of heat treatment facts, figures, and terminology.

The "quench-hardening" of steels means dissolving the compound by heating to the solutioning temperature indicated by the diagram, and then cooling at a rapid rate *in the attempt* to retain the solutioned state for treatment as convenient by tempering later on; it is called "quench-hardening" because the original steels always hardened on quenching in water; the solutioned structure is not trapped by quenching, or else is incompletely trapped; therefore, all or much of the compound is reborn from the solution in a molecular dispersion which produces a hardening.

Though vital from practical standpoints, this hardening is incidental and trifling from the standpoint of theory; it simply means that solutioning has gone forward properly, but the quenching has not been altogether successful in holding what the solutioning has attained, having allowed something to slip past and blunder into the initial stage of the next process—tempering.

Now, the "solutioning and precipitation" of the non-ferrous heat treatment is perhaps a more rational and understandable phrase, but it should be remembered that "solutioning" means dissolving and *quenching as well*, which latter is nearly always successful in holding what the solutioning attains. The "precipitation" in the non-ferrous is the "tempering" in the ferrous, and the effects are always procured by identical means; methods may seem strangely different in some cases, but a proper understanding of such oddities as age hardening and temper hardening will prove that the methods are the same in all cases.

Age hardening is precipitation which takes place spontaneously in the case of some alloys; this simply means that for such alloys room or ordinary temperatures are tempering or precipitation temperatures. Keep them at a sufficiently low temperature and they will remain solutioned indefinitely. Temper hardening seems odd because tempering usually softens; it always softens a *thoroughly* quench-hardened alloy by inducing the softer transition products; but if quenching does not harden, or does not completely harden, the solutioned alloy has been entirely or partly retained by quenching, and tempering must in these cases produce hardness (or extra hardness) *first*—corresponding to martensite formation—followed by softening as usual. Hence tempering or precipitation by no means necessarily produce softness at the lower tempering temperatures, though they may do so at the higher.

The Steel part of the table will be better grasped if it is remembered that steel is described as, or called, whatever its microscopic structure happens to be on cooling in air "normally" from its austenitic condition; whatever its alloying elements, a steel is "pearlitic," "sorbitic," etc., if on cooling from above its critical range in air it appears to have a pearlitic, sorbitic, etc., arrangement of its constituents or components. This explanation accounts for the fact that certain forms of the steels are not obtainable from other forms except under conditions of cooling *slower* than cooling in air, or else on taking down to *low temperatures* at the air-cooling rate.

Specific cooling rate and specific final temperature are factors which work in conjunction and in the same sense to procure the different forms or structures of steels; high cooling rate and high final temperature will each procure the higher transition products or austenite; low cooling rate and low final temperature will yield the lower transition products and pearlite or even spheroidized pearlite.

Solutioning and quenching in the case of Duralumin heat treatment is commonly known as "normalizing": it is a pity that words which could be useful terms in a proper terminology are so abused, but it is necessary to know what is meant nowadays by each word in common use whether it is a good word or a bad one.

The heat-treatable compositions of some of the non-ferrous alloys should be distinguished carefully from the alpha, or simple solid solution compositions of the same alloys not intended for heat treatment of this kind. Aluminium bronze simple solid solution alloy has but 5 to 7 per cent of aluminium; tin bronze simple solid solution has but 10 per cent of tin; the object of the founder in each case being to keep sufficiently below the composition which yields the compound to ensure that no compound will appear even on the rapid cooling of these alloys. (Chapter X, page 57, on "Coring.")

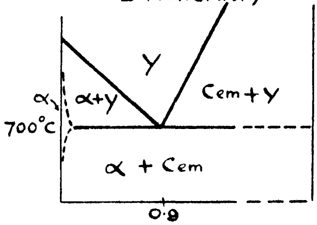
In the concluding example of this kind of heat treatment, the

ANALOGOUS CONSTITUTION OF HEAT-TREATABLE ALLOYS

(ROUGH SIMPLIFIED DIAGRAMS)

THE CARBON STEEL DIAGRAM

(PART OF THE IRON-CARBON DIAGRAM)



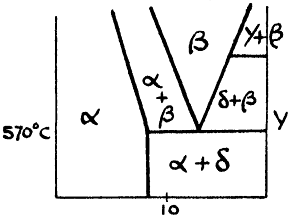
CARBON PERCENTAGE

(A RANGE OF CARBON PERCENTAGES IS UTILISED FOR HEAT TREATMENT)

N.B THERE IS ACTUALLY A VERY SMALL α SOLID SOLUTION LIKE THE OTHER DIAGRAMS

THE ALUMINIUM-BRONZE DIAGRAM

(PART OF THE COPPER-ALUMINIUM DIAGRAM.)

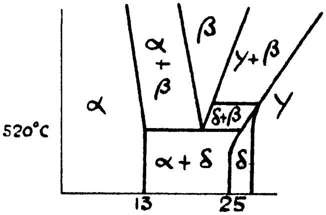


ALUMINIUM PERCENT.

(ONLY THE 10% ALLOY IS UTILISED FOR HEAT TREATMENT)

THE TIN BRONZE DIAGRAM

(PART OF THE COPPER-TIN DIAGRAM)

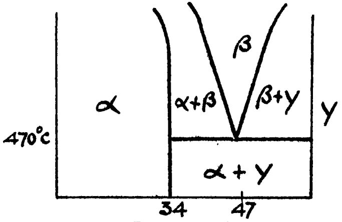


TIN PERCENTAGE

(13 TO 25% TIN ALLOYS MAY BE UTILISED FOR HEAT TREATMENT)

THE BRASSES DIAGRAM

(PART OF THE COPPER-ZINC DIAGRAM.)



ZINC PERCENTAGE

(HEAT TREATMENT GENERALLY RESTRICTED TO WORKING α/β AND β BRASSES ABOVE 470°C. BELOW WHICH THE EUTECTOID GIVES HARDNESS AND BRITTLENESS.)

N.B. THIS IS SOMETIMES DESCRIBED AS β', ANOTHER SOLID SOLUTION DIFFERENT FROM β.

FIG. 29

high speed tool steel, a more complex state of affairs exists which is quite easily analysed.

To begin with, the steel is hyper-eutectoid containing an excess of alloy carbide which is not completely solutioned; this excess forms granules which remain embedded in the final alloy for use as a cutting steel; as solutioned, the steel is an alloy austenite; as quenched in oil or in air the steel is still mainly austenite, but now has a stiffening network of martensite among the austenitic grains; tempering at low temperatures only softens the martensite; temper hardening at higher temperatures hardens the austenite by inducing molecular dispersions of two different kinds of compound—iron tungstide and the alloy carbides—both of which procure a martensite formation, progressively as tempering or using temperatures rise, with progressively added strength and hardness to counteract the increasing softening of the original martensite, to counteract also the increasing softening which high temperatures commonly induce in all metals.

It should be noted that, whereas in the case of steels (iron-base alloys) there are a great number of highly important combinations of carbon and of the alloying elements which require fairly complete knowledge of the carbon steel diagram and of the various alloy steel diagrams, in the case of the non-ferrous heat-treatable alloys their numbers are extremely few at present owing to the newness of the discovery that non-ferrous alloys could be heat treated. In any one group, such as the copper-base alloys, it may be more reasonable to turn to a list of a few solutioning and tempering temperatures known to give satisfactory results, than to make use of the copper-aluminium diagram or the copper-tin diagram.

Pearlite owes its name to the mother of pearl sheen of the form containing its carbide in layers or laminations, only obtained by the cooling of austenite. When a eutectoid structure of a *similar degree* of coarseness of carbide is developed by the tempering of quench-trapped austenite or martensite, these laminae are not formed, and the product does not show a pearly lustre.

There is little point, however, in giving this latter structure a completely different name, and "granular pearlite" is the name sometimes given it. For, what matters so much more than the mere play of colours is that the degree of cementite dispersion in such a product of tempering can be made the same as that of the laminated form, after which its mechanical properties are found to be comparable with those of the laminated form.

CHAPTER XXI

MICROSCOPIC STRUCTURES AND PROPERTIES

FOR the sake of engineers whose particular interest is in the mechanical properties, a particular aspect of metallurgy has been presented with rather unusual relentlessness. This has led to the somewhat strange conclusion that properties are modified almost entirely by sub-microscopic or lattice effects; yet the knowledge that microscopic evidence is of great value may lead to misgivings with regard to the explanations here set forth.

If the position is rightly understood, no contradiction is involved in the discovery that sub-microscopic invisible machinery is at work in determining all the really important properties and property changes. The evidence of the microscope is invaluable—not, however, for showing up structures which we can recognize *per se* to be strong, hard, ductile, or tough, but for showing up that evidence of the metal's state or condition which "pin-points" the sub-microscopic structure enabling estimates of the property-forming machinery to be made with the aid of X-rays.

Even the above restatement of the position is not quite satisfying, and a rather more careful analysis of the question would not be out of place—

When we are considering various electrical properties of the metals, it is rare to find that either the visible structures or the invisible ones can be construed usefully to mean anything of the electrical character sought: researches aimed at endowing metals with the electrical properties are at present like fumblings in the dark with the valuable evidence of previous experience only as a guide.

An extremely opposite case to this is the design of a bearing metal: if certain somewhat coarse mechanical properties are desired, such as a hard component set in a soft matrix for reasons connected with the lubrication of a shaft, to a large extent the microscopic examination shows just what is necessary, the distribution of components of required size and hardness equally throughout the ground-mass of the required softness, strength, and ductility. There seems no need at all to worry with the space lattice. In a bearing, if you know the mechanical properties of both the hard and the soft component, you can feel that the two are "pulling together" in the structure which is clearly visible; you can actually see the required mechanism under the microscope.

But in the structural alloys with which engineers are chiefly concerned the position is vastly different. There are certainly a few alloys which, being but fairly coarse mixtures of two components with well-known properties, seem understandable enough without any ramblings into lattice questions. Even if it is possible to explain the difference in properties between pure iron and coarse pearlite (a matter of 20 tons per sq. in.), on the assumption that pearlite is but a mixture of iron and carbide of iron with a structure which is "obviously strong"(?), it is utterly impossible to account for a difference of properties (amounting to nearly 200 tons per sq. in.) between pure iron and the hardened alloy steels in any such manner.

The law with regard to mixtures—that the properties are the arithmetical averages of the properties of the components—is applicable to alloys only with the greatest of caution, and it rarely fits even in the case of coarse mixtures of alloy constituents in the "as cast" condition. So much depends on the deeper structures at the crystal boundaries, and this point raises the questions of boundaries and grain size—for the smaller the grain, the more the boundaries.

Grain Size

A large range of grain size can be seen under the microscope at various magnifying powers, perhaps the whole range of any practical importance. To what extent does grain size affect the mechanical properties? Is it not possible to understand these property changes without dragging in sub-microscopic structures?

The most striking change of property with change of grain size is that of notch-toughness (resistance to the propagation of a crack once formed). Notch-toughness rises considerably as grain size falls. Can this be visualized readily? Strength, hardness, and ductility are all slightly improved by fall in grain size. Is this reasonable? Creep resistance falls with grain size. Can we make anything of this without turning the sub-microscopic structures?

The usual polycrystalline metal to which the foregoing applies fractures chiefly across the grains, but partly also along the more directly interposed boundaries. It might be said that the more finely jagged fracture path of the metal of finer grain would obviously resist the propagation of a crack already started; it might even be said that the more finely jagged fracture path would surely be the longer one, but the nature of the actual fracture path can have *nothing to do with strength or ductility, because these register themselves in their tests before fracture itself commences.* These are all more likely to have been improved because of the increased amount of the sub-microscopic "amorphous" material of the boundaries which is contained in the finer grained metal; and creep resistance is just the property which you would expect to fall in the presence of an extra amount of amorphous material.

So grain-size effects, though perhaps partly explainable by observation direct on the basis of a jagged or lengthy fracture path, cannot be properly dealt with unless sub-microscopic considerations of the properties of the "amorphous" metal in the grain boundary are taken into account.

Steel Refining and Normalizing

Returning to steels, straight carbon steels between the very mildest and the very highest in carbon differ in strength by about 20 tons per sq. in.—in the annealed state, that is. Such steels are mixtures of iron and iron carbide; the increase in strength rises to the pearlitic structure and then falls again. It is fairly easy to imagine from the microscopic view that the mixture forms a strong structure up to the pearlitic 0·9 per cent of carbon, and that beyond this the mixture structure is not so good.

Perhaps this is the case; but even if this is admitted, the property change from 0·1 per cent C. to 1·7 per cent C. is comparatively small. No such explanation on self-evident lines can be invoked to account for the changes taking place in iron to give it the 200 tons per sq. in. U.T.S. of a hardened special steel, but there is every reason for supposing that, with the aid of X-ray evidence, sane conclusions can be arrived at to account for these enormous changes *on the lattice theory.*

In fact it is found that, even as microscopic structures get finer, property changes become more marked; compared with an annealed steel, the refined steel is an improvement, but there are even greater changes in properties when a refined steel is converted into a lamella sorbitic steel, with the same structure only finer still.

The whole business of quenching and tempering is, in effect, aimed at getting the carbide texture in iron finer and still finer, until it attains the greatest possible refinement—that of molecular dispersion—which gives rise to the most acute mechanical properties of all in martensite; this dispersion is bound to be quite invisible under the microscope.

Similarly, no visible change takes place when Duralumin age-hardens.

Flow Structure

Flow structure is visible enough, and is very rightly said to be of importance, warning us to make use of the metal in which it exists

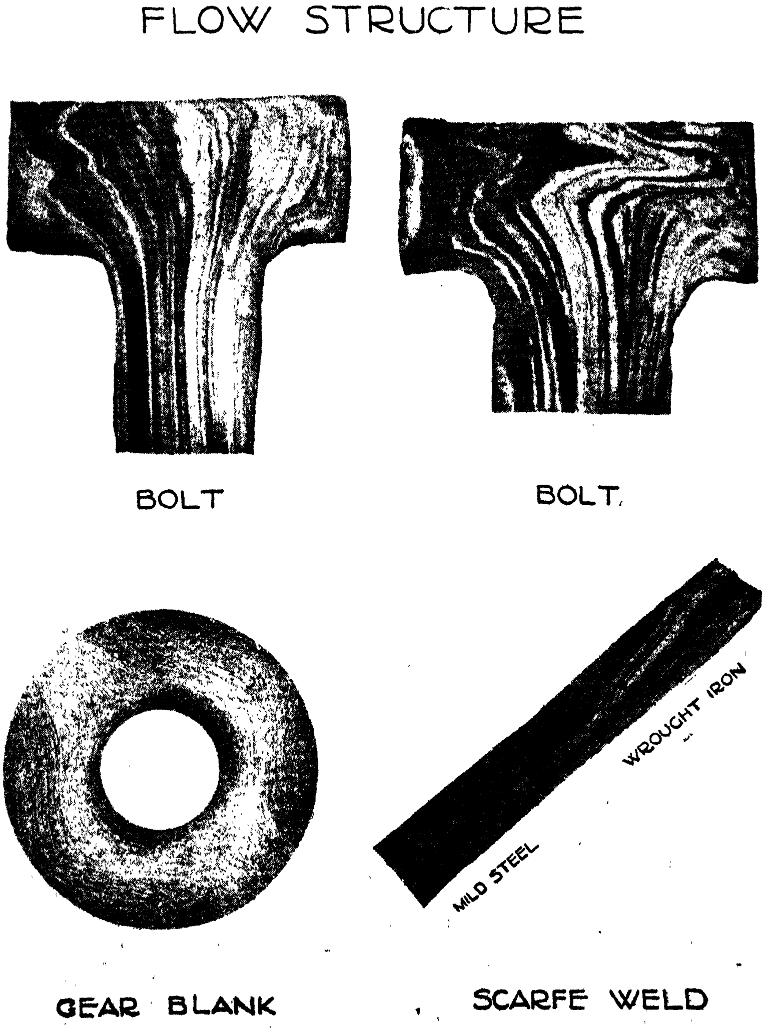


Fig. 31

so that stresses of tension and compression are laid along the flow, while stresses of shear are across the flow.

In fact, wherever weak constituents or impurities, or cavities or regions of porosity have long axes and are orientated in the mass there is some danger that a path of weakness may be traced throughout the metallic mass by them, or else that they may act as local stress raisers under shock conditions.

Weak constituents, etc., which are scarce, evenly distributed, and equiaxed give little trouble unless they are coarse by comparison with the two-inch test piece on which the properties are tested. They may still act, however, as highly dangerous stress-raisers to lower the Fatigue Limit of the metal, even if they are extremely fine, if they happen to be of the special *fatigue* stress-raiser type.

CHAPTER XXII

ALLOY SPECIFICATIONS AND MARKINGS; TEMPERING CHARTS

ALLOYS are registered and stamped by the firms producing them with certain private or proprietary names, letters and numbers—such as "Elektron Alloy AZ.31" (a magnesium-aluminium-zinc-manganese light casting alloy)—sometimes called their "civil designations." In this way the same alloy will be given different designations by different producing firms. Proprietary colour markings are also employed.

The Department of Technical Development of the Air Ministry takes into use and registers a large number of alloys for Royal Air Force purposes, giving these "D.T.D." specification numbers such as "D.T.D. 59" for example, which for all intents and purposes is the same as the above AZ.31; it may be said that Elektron Alloy AZ.31 meets the requirements of D.T.D. 59.

Important alloys which have become thoroughly standardized as a result of well-proved usefulness in engineering are also registered and recognized by the British Standards Institution, and are then given "B.S." letters and numbers.

The "Aircraft Series" of these are the more important, but there is also the "General Engineering Series," distinguishable by the abbreviation "No." preceding the number, which is often followed by the year of issue.

Special colour markings for the different B.S. and D.T.D. specifications have been introduced by the Aeronautical Inspection Directorate.

A certain light alloy suitable for pistons and air-cooled cylinder heads, prepared according to the formula of Messrs. Rolls Royce Ltd.'s Hiduminium Series, is known to that firm and fairly generally as "RR.53." Adopted by the Air Ministry it becomes known also as D.T.D. 131 (D.T.D. 238 is the same alloy "as cast"). Of the B.S. specifications L.35 is the most similar, but is infrequently used nowadays.

To consider a few of the specifications for aircraft and engine construction alone, different specifications are needed, in *steels* for example, to indicate and distinguish—

(*a*) Forms of supply, e.g. bars, forgings and sheet; tube; wire and cable; nuts, bolts and test pieces.

(*b*) Conditions of supply, e.g. bright, normalized, heat-treated.

(*c*) Varying strength and toughness in carbon and low alloy steels.

(*d*) Corrosion resistance (low, medium and high tensile).

(*e*) Carburizing case-hardening (different tensiles).

(*f*) Nitriding case-hardening (different tensiles).

(*g*) Resistance to heat.

(*h*) Requirements as to coefficient of expansion by heat; and other identifying considerations.

Examples of these are as follows—

(*a*) "S" stands for bars, forgings and sheet; "T" for tubes; "W" for wire and cable; "A" for bolts, nuts and test pieces.

(*b*) Steels B.S. 3S.6 and 2S.76 are identical in composition and in uses, but the first is a supply in normalized condition, and the second in the oil-hardened and tempered condition. Their properties differ.

(*c*) (i) S.1 is supplied in "bright" bars for lightly stressed nuts, bolts and machined fittings in general.
  (ii) S.6. for medium tensile fittings.
  (iii) S.11 for high tensile bolts, studs, fork-ends, sockets.

(*d*) (i) S.61 is a low tensile "stainless iron" for lightly stressed machined fittings.
  (ii) S.62 is a higher tensile stainless steel.
  (iii) S.80 is an incorrodible steel such as is used in fittings subjected to severe corrosion and high stress (as in flying-boats).

(*e*) (i) S.14 is a mild carburizing case-hardening steel for gudgeon pins and camshafts.
  (ii) S.15 is a stronger 3 per cent nickel case-hardening steel for the above purposes, and also for gears, tappet rollers and tappets.
  (iii) S.90 is a 5 per cent nickel extra high tensile case-hardening steel used in airscrew reduction gears, gun gearcams, and for all the above purposes as well.

(*f*) (i) D.T.D.228 is a nitriding steel for crankshafts and other hard wearing parts.
  (ii) D.T.D.306 for the same rather more highly stressed parts.

SPECIFICATION CHART
OF A FEW WELL KNOWN STEELS.

SPECIFICATION Nº	COMPOSITION.								MECHANICAL PROPERTIES							FORM AND CONDITION IN WHICH SUPPLIED	TYPICAL USES.
	CARBON %	SILICON %	MANGANESE %	NICKEL %	CHROMIUM %	MOLYBDENUM %	VANADIUM %	TUNGSTEN %	MAX STRESS TONS/□"	PROOF STRESS TONS/□"	ELONGATION %	RED. OF AREA %	IZOD IMPACT FT. LBS	BRINELL Nº		
S.1.	0·15-·40	0·30 MAX	0·5-0·8	–	–	–	–	–	35-45	–	15 MIN.	40 MIN.	–	–	Bright bars-cold rolled drawn or machined	Nuts bolts and lightly stressed machined fittings.
S.6.	0·35-·45	0·30 MAX	1·20 MAX	1·0 MAX.	–	–	–	–	35-45	–	20 MIN	–	20 MIN	146-201.	Bars, forgings, stampings – normalised	Cylinder barrels, fuel pump barrels, (engine). Bolts, nuts and machined fittings, (airframe).
S.76.	0·35-·45	0·30 MAX	1·20 MAX	1·0 MAX					40-50	–	22 MIN	–	35 MIN	174 223	Bars, forgings stampings, – oil-hardened and tempered.	
S.11.	0·25-·35	0·30 MAX	0·45-·70	2·75-3·75	0·50-1·0	0·65 MAX	0·25 MAX	1·0 MAX	55-65	–	18 MIN	–	40 MIN	248 293	Bars, forgings, stampings, – oil-hardened and tempered	Engine – Airscrew shaft and hub, crankshaft, stressed bolts, misc., nuts; Airframe. – Nuts, stressed bolts, fork-ends, sockets, tube-plug ends, wing roots, eyebolts, undercarriage mounting fittings.
S.14.	0·10-·18	0·30 MAX	0·90 MAX	–	–	–	–	–	32 MIN	–	20 MIN	50 MIN	40 MIN	–	Bars, forgings, stampings.	Valve tappets and rockers, rocker rollers, camshafts, fuel pump gears, water pump spindles.
S.15	0·10-·15	0·30 MAX	0·20-·60	2·75 3·5	0·30 MAX	–	–	–	45-60	–	18 MIN	45 MIN	40 MIN	–	Bars, forgings, stampings.	Valve tappets, tappet rollers, push rod heads, timing wheels, pinions.
S.61	0·15 MAX	0·50 MAX	–	1·0 MAX	12·0 MIN	–	–	–	35-45	–	25 MIN	50 MIN	45 MIN upto 2" dia. 25 MIN over 2" dia.	152 207	Bars, forgings, stampings, – oil-hardened, and tempered	Engine – Throttle control rods, and pins, and control elbows; Airframe – Lightly stressed machine fittings.
S.62.	0·15-·65	0·50 MAX	–	1·0 MAX	12·0 MIN	–	–	–	45-52	–	20 MIN	45 MIN.	35 MIN upto 2" dia	207 235	Bars, forgings, stampings, – oil hardened, and tempered.	Engine – Inlet valves, throttle control rods, and pins, pump spindles, carburetter parts, and exhaust studs. Airframe. – Bolts and nuts, wing fittings, fork ends, and eyebolts in the end frame.
S.80	0·25 MAX	0·50 MAX	1·0 MAX	1·0 MIN	16-20	–	–	–	55 MIN	–	15 MIN	–	25 MIN	241 MIN	Bars, forgings, stampings, – air-hardened and tempered.	Bolts and nuts, fork ends and pins, sockets, tube end plugs, wing roof fittings, eye bolts, wing fixing lugs, under-carriage, and engine mounting fittings, amphibian and retractable undercarriage parts, turnbuckles, levers, chain parts.
S.90	0·16 MAX	0·30 MAX	0·60 MAX	4·50-5·0	0·30 MAX	0·50 MAX	0·25 MAX	1·0 MAX	65 MIN	–	13 MIN	–	30 MIN	–	Bars, forgings, stampings.	Airscrew reduction gears, cam gears, camshafts, gun gear cams, gudgeon pins, magneto drive wheels.
D.T.D 228	0·25-·35	0·35 MAX	1·0 MAX	1·0 MAX	0·5-1·5	0·9 1·5	0·25 MAX	1·0 MAX	55-65	–	18 MIN	–	35 MIN	248 293.	Bars, forgings, stampings, oil-hardened and tempered.	Crankshafts, airscrew shafts, and other hard-wearing parts of the engine.
D.T.D 306	0·15-·35	0·35 MAX	0·65 MAX	0·30 MAX	2·5 3·5	0·30 0·70	0·25 MAX	1·0 MAX	60-70	–	17 MIN	35 MIN upto 2½ dia 25 MIN over 2½ dia		269 321	Bars, forgings, stampings, oil-hardened, and tempered	Master and articulated connecting rods, crankshafts and airscrew shafts.
D.T.D. 49B.	0·35-·50	1·0-2·5	1·5 MAX	10·0 MIN	12·0-16·0	–	2·0-4·0	–	–	–	–	–	15 MIN	269 MAX	FORGINGS, SOFTENED	VALVES.

FIG. 32

(*g*) D.T.D.49 for the heat resistance of inlet and exhaust valves.

(*h*) D.T.D.247 for valve seats and for studs holding down high-expansion light alloy cylinder heads.

Figures standing before the letter of the B.S. specification merely refer to the number of times the same specification has been reissued, and can usually be ignored as the latest issue is always referred to. In the D.T.D. series, however, the issue is denoted by means of letters A, B, etc., *following* the number. These particular letters of the D.T.D. series must not be confused with A, B, C and E following the number in the B.S. series, for the latter refer: "A" to bars or billets for forging; "B" to the same heat treated for machining; "C" to forgings themselves; "E" to special forgings of aero crankshafts and airscrew shafts. A, B and C in B.S. sheet and strip refer, on the other hand, to: "A" softened sheet; "B" softened strip; and "C" heat treated or cold rolled sheet or strip. In the D.T.D. series "Sections II, III, IV and VI" take the place of the A, B, C and E following the number in the B.S. specifications.

Aluminium Metal and its Alloys—Light Alloys

B.S. specifications of these are given the letter "L" and a number. L.33 is "silicon alloy"; L.1 bars up to 3 in. and small forgings, bars over 3 in. and larger forgings are covered by L.39—both in Duralumin; the same in sheet Duralumin is L.3, in rivets L.37, but in tube "T.4" is the specification. L.35 is the old Y alloy, now hardly used in aero work except for the Bristol cylinder head. Airscrew forgings and general purpose wrought Duralumin have D.T.D. numbers 147 and 150; as one of the Hiduminium Series it is known as "DU Brand."

The Magnesium Alloys are known by their civil designations for the most part, as there are not many D.T.D. specifications and there are no B.S. specifications.

Copper, the Brasses, Bronzes and Bearing Metals are given the letter "B" in B.S. aircraft specifications. "B.S.S." belongs to copper and copper alloys of the General Engineering Series.

The General Use of the Specifications

Owing to the number and variety of all these unsystematic and uncorrelated specification schemes, identification of raw metallic material supplied cannot as a rule be made from memory, so that reference to full specification charts becomes inevitable. Relatively few of these designations can conveniently be committed to memory, and therefore it seems somewhat unsatisfactory to use any of them for the sake of brevity.

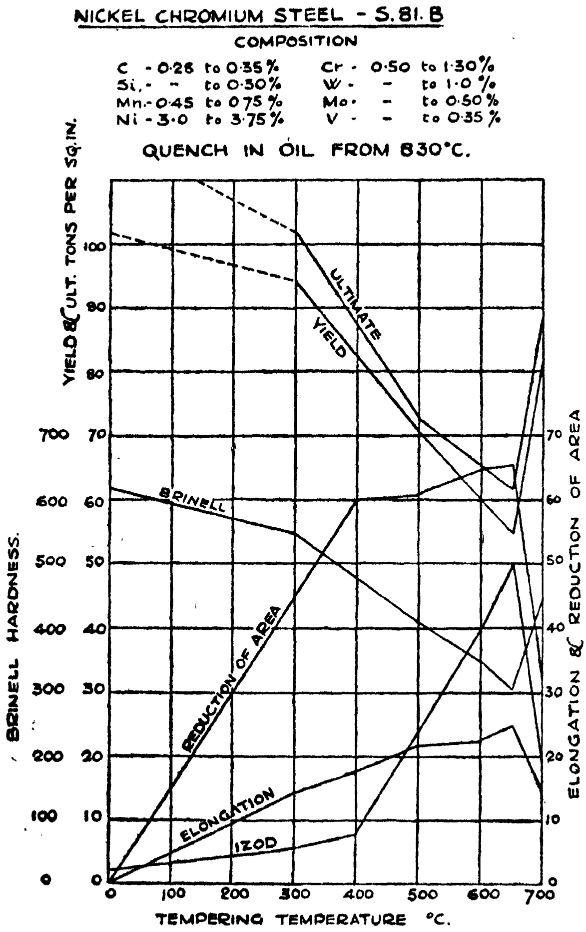


FIG. 33

Very different would be the case, however, if it were ever found possible to arrange a code wherein each letter and number stood for an essential in the metal or alloy to identify it as closely as necessary. Organizers of complex filing systems, and chemists also, have had to face similar problems. It would be convenient and desirable to make constant use of specification letters and numbers which were full of meaning to all who had taken a little trouble to learn the code.

Specification Charts are tables of specifications against each of which is shown the—

(i) Composition.
(ii) Mechanical Properties.
(iii) Form and Condition in which supplied.
(iv) Typical Uses.

(Sections A, B, C and E and colour markings are also included in a full chart.)

A few examples from a common type of specification chart are given in a table.

A Tempering or Heat Treatment Chart for each steel or other alloy is a more detailed way of recording and of showing at a glance how the mechanical properties vary with variation in heat treatment.

An example of such a Tempering Chart is given; and from this it will easily be seen that, after "quench hardening" (solutioning and quenching) steel S.81, tempering and quenching at progressively higher temperatures up to 650° C. progressively reduces strength and hardness and increases ductility. These changes take place because the martensite trapped by the solutioning and quenching is being progressively let down to pearlite by tempering right up to a tempering temperature of 650° C.

Tempering beyond this temperature has the opposite effect, the reason being that 650° C. is the highest temperature for tempering *followed by quenching* which will give the pearlitic structure: higher tempering followed by quenching brings in martensite once again very rapidly increasing in amount as the tempering temperature rises.

The properties of S.81 call for a tempering treatment between 500° C. and 650° C.; but, apart from this, a steel of the composition there shown could naturally be provided at will with various properties by the simple process of examining the Tempering Chart and choosing whatever tempering temperature suited the particular object in view.

CHAPTER XXIII

CORROSION

1. The "Mechanism" of Corrosion

Corrosion is destruction by chemical or electro-chemical means; in some cases chemical agencies (such as chlorine or oxygen) can proceed without electrical agencies when the superficial product of the reaction is soluble in or penetrated by the reagent; but electrical agencies almost invariably assist chemical agencies, and are very active under practical conditions; they originate from—

(i) Potential differences peculiar to dissimilar metals in contact (the true basis of the differences of potentials of the "electro-chemical series"—see below).

(ii) Local P.D.s between parts of the same metal which happen to be either in different states of plastic strain, or in different states of stress.

(iii) P.D.s locally at a metal surface in an electrolyte owing to differences in the concentration of that electrolyte in contact with a homogeneous metal.

(iv) P.D.s due to differences in the aeration of different parts of a homogeneous metal when, by local excess of oxidation, part of the metal behaves as a dissimilar metal (i.e. as a more electro-positive metal).

Wherever these differences of electrical potential occur, and when at the same time the electric circuit is completed by an electrolyte, a current can flow; the electro-negative part will dissolve in the liquid, hydrogen gas being deposited at (or given off from) the electro-positive part. Such a current may continue to flow until all the water of the electrolyte is finished, or until all the dissolving part has been corroded away.

An **electrolyte** is readily obtained under natural or commonplace conditions such as by the solution of acid atmospheric gases in water. Acid moisture, and sea or other salt waters deposited on a surface carrying any of the above P.D.s, will give rise to corrosion.

Atmospheric oxygen may completely and imperviously coat a metal with a protective film resisting penetration by the electrolyte, or else it may differentially aerate the surface of the metal giving rise to corrosion according to (iv) above.

2. The Property of Resisting Corrosion

A perfectly non-reactive element cannot corrode, but many metals appear to be non-reactive under special conditions on account of the formation of a film.

The "passive" state of certain metals has usually been found to be due to the oxide film given either by atmospheric oxygen or by the oxygen of nitric acid; with such a coating these metals do not react further with the oxidizing agent, and sometimes they become unreactive to certain other reagents as well.

Passivity in General.—The insolubility of a metal in an acid or other reagent is as a rule on account of the insolubility of the product which would be formed if the reaction were to go forward; the suggestion therefore is that the reaction does go forward to the formation of an insoluble film of the product on the surface of the metal causing the apparent lack of reactivity, or the insolubility of the metal. Plenty of evidence supports this view.

(It would appear, therefore, that one universal way of preventing corrosion would be to contrive that there were always present some element in the alloy to produce a reagent in the electrolyte capable of forming an insoluble product under the conditions of the corrosion, e.g. a silicate—a proposal more easily made than carried out.)

The evident solid products of corrosion, such as rust or verdigris, cannot be expected to give protection against further corrosion because, unlike the films giving passivity and protection, these products are not formed *on* the surface of the metal, but within the electrolyte at some distance from that surface. Far from giving protection, these solid products will usually give rise to differential aeration, and will thus be the cause of still further corrosion.

3. Anti-Corrosion Treatment

Anodizing.—Aluminium is notably the metal which takes a heavy protective coating when it is made the anode in an electrolyte cell, oxygen ions being discharged at the anode; others like nickel, iron, and copper may be anodized but without so much lasting advantage.

Protective Metallic and other Coatings. Some of these are provided because they are electro-negative to the protected metal, and are therefore attacked by the corrosive agents in preference to the metals they coat as soon as the electrolyte reaches the contact between them, e.g. zinc on iron. Others, while electro-positive to the coated metal, are nevertheless more resistant themselves on account of their own passive oxide films, e.g. tin on iron, and therefore last a long time as coatings; wherever they are penetrated, however, corrosion may be deep and extensive.

Coatings may be applied by—

(i) Dipping into molten metal, as in tinning or "galvanizing" iron.

(ii) Electroplating, as for cadmium, chromium, nickel, silver and gold coatings.

(iii) Spraying with liquid metals (of almost any kind).

(iv) Mechanically, by rolling coating sheets on to a metal slab; or by drawing or rolling a composite ingot obtained by casting the protecting metal around the metal to be protected—as in the production of Sheffield Plate, gold "filled" articles, or Alclad and similar composite metals.

(v) Cementation processes, whereby the metal surface is superficially alloyed by heating with a powder of the protecting metal, as in—

Sherardizing of steels with Zinc.
Chromizing of steels with Chromium.
Calorizing of steels with Aluminium.
(Nitriding of steels with Nitrogen of Ammonia Gas.)

(vi) Coslettizing or Parkerizing with a phosphate film (with or without additional enamel coating).

(vii) Paints, varnishes, dopes, enamels, and so forth, either alone or in conjunction with one of the above coatings.

Normal Potentials

(*The Electro-chemical Series of Common Metals*)

(Hydrogen = 0·0000 volts)

Platinum	+ 0·86	volts	Cobalt	– 0·23	volts
Silver	+ 0·7987	„	Iron (ferrous)	– 0·42	„
Mercury (Hg)$_2$	+ 0·7928	„	Cadmium	– 0·44	„
Copper	+ 0·3469	„	Zinc	– 0·77	„
Hydrogen	0·0000	„	Aluminium	– 1·337	„
Lead	– 0·132	„	Magnesium	– 1·8	„
Tin	– 0·146	„	Sodium	– 2·715	„
Nickel	– 0·20	„	Potassium	– 2·925	„

CHAPTER XXIV

CURATIVE MEASURES IN GENERAL

(in the case of damage to alloys)

THIS account has dealt mainly with the deliberate methods of modifying properties when the initiative has been in the hands of the metallurgist; some of the results of that initiative, the common alloys, have been discussed.

Accident or Damage

The other side of the picture, the initiative having passed out of the hands of the metallurgist, is worthy of a little more attention. It is worth while envisaging the possibility of the break-down, from one cause or another, of all such deliberately planned treatments; it is desirable to take note of the effect on metals of accidental fire, of occasional gross overload, of fatigue in a burnt or corroded state; and it is essential to be able to assess the possibilities of repair in case of severe mechanical damage involving plastic distortion and/or fracture.

The subject now being touched upon in some of its more urgent aspects is, of course, a very large one beyond the scope of this Course; but, although the following may show how complex the subject is, it is by no means indicated that the engineer will be obliged to discard all metal which has suffered in an accident of some kind. The object of this chapter is rather to show that *the less an engineer has taken his essential metallurgy to heart, the less ought he to risk using any doubtful metal, and therefore the more finished metal must he waste.*

Heat treatment, upset by the failure of gas or electric power, by breakdown of furnace protective atmosphere, by inaccuracy of pyrometers, or in any other way, can usually be recommenced and carried right through to a successful conclusion *unless* the incident has somehow involved the metal's exposure to too high a normalizing or solutioning temperature for a long time. The effect of this is to coarsen the grain of steels, and to destroy the uniformity of distribution of components in the case of other alloys. The forerunner and even the beginning of fusion may take place with the latter in the form of segregations of components into the grain boundaries and corners (whereat molten eutectics ultimately assemble) to destroy the homogeneity and uniformity of composition. In this way Duralumin, Y alloy, the similar but improved Hiduminium alloys, and the high speed tool steels have but narrow margins between solutioning and fusion temperatures: exceeding the former substitutes a most undesirable embrittlement for the beneficial effects of the correct heat treatment.

How then, if *incipient* fusion is so bad, can satisfactory heat treatment ever be carried out after the total fusion and casting which all alloys must undergo? The answer is to be found in the very reason itself for "chill-cast" and "wrought" alloys. An alloy containing the percentage of components requisite for suitable heat treatment effects may segregate badly if given the chance to cool slowly as in the case of the metal in a large ingot mould or a sand mould; chill-casting into the final shape, or chill casting an ingot for subsequent hot working into final shape by extrusion, forging, pressing, stamping or drawing may be essential to distribute the components properly; and the latter treatment may also be of the greatest value for the production of a fine grain and the flow-structure suitable for some particular metal member of an engine or airframe for example. The heat treatment by normalizing or for optimum properties has, therefore, to be *superimposed* upon these essential preliminary treatments; and the undoing of these latter by overheating may create an insoluble problem for the engineer because he may find himself unable to re-distribute the components and reinstate the desirable flow structure, though he may be able to restore the fine grain. The coarse grain in steel is corrected by normalizing once or several times.

Such alloys have, therefore, been planned in composition for satisfactory heat treatment *after* a sufficiency of chill casting and/or hot working; they are some of the more subtle alloys which must be understood from the angle of the metallurgist, and cannot by any possibility be handled until that angle is appreciated.

Upon the properties of such alloys the lives of flyers and others often depend, so that it is perhaps relevant to draw attention to the remarks of the Introduction on the responsibility of engineers. In fact, it is of the greatest importance before doing anything with such alloys *first of all to ascertain as nearly as possible, and to understand sufficiently, how the metal part in question came by its final state* before it was built into the engine or structure in question.

Accidental Fire

Examination and Test. It may be quite impossible to guess the effect of exposure to a high temperature flame; the highest temperature may only have been reached at a mere skin, good conductivity having saved the bulk of the metal from dire consequences. An obviously coarse crystalline exterior would be ominous, but indentation tests and microscopic examination on hacksawn flat portions of the metal should be allowed to bear witness to the condition of the bulk if this is in any way possible. Also, if possible, a section in depth should be taken to ascertain how far the damage has gone; and a small tensile test piece or two would help greatly, though it is not often such flatting or such a cut can be made into a metal part which might be saved.

Exposure to fire or heat may result in—

(*a*) The tempering or even the full annealing of a heat-treatable alloy, calling for a full heat treatment for the restoration of original properties.

(*b*) Local solutioning, not reaching the dangerous incipient fusion above described: a new heat treatment is called for.

(*c*) Local incipient fusion, and the above-described dangers of damage to distribution of components and to flow-structure. The original "high duty" properties may not be wholly restored on heat treatment, but a good approximation to them may be obtained by repeated heat treatments. (The assumption is made that it will be impossible to imitate the original hot work procuring the original grain and flow-structure.)

(*d*) Superficial burning or penetration of the boundaries by oxygen and nitrogen, whereby fatigue and shock resistances will suffer unless such a skin can be removed with impunity.

(*e*) Decarburization of steels in an oxidizing flame, or in some quenching oils; or carburization in a reducing (smoky) oil or gas flame: unless the metal is a special heat-resisting steel, thin sections, tubes, and wires may suffer serious deterioration.

(*f*) Extension, flexure or other plastic distortion under stresses beyond the elastic limit of the hot metal, causing enlargement of bolt and rivet holes, sagging, malalignment, etc., and a new distribution of stresses on cooling not reckoned for in the design. This is an extreme case when the worst has happened, and yet it may be imperative to effect the best repair possible. Reheating to hot-working temperatures for straightening, etc., would always be the preliminary; this would be followed by any necessary welding or building up by welding; normalizing or any other necessary heat treatment would now follow, and repetitions of these might be needed to produce a really good result; finally, all the finishing work could be done, the original surface removed as much as permissible and the new surface left bright.

Gross Overload

Loading beyond the safety factor may prove, upon examination, to be harmless enough; but loading beyond the elastic limit will result in at least some change of dimensions. Much of the following can only apply to cases of overloading in which change of dimensions is not a critical consideration.

Step by step with changes due to mild cold work there will be induced a certain strengthening and hardening; loss of ductility and of toughness will not be serious, but internal stresses may easily be of dangerous height.

The fullest amount of plastic strain right up to fracture point may escape attention if it is caused by pure alternating stresses; no visible trace can ever be seen of fatigue damage except a dulling of an otherwise bright surface until the fatigue crack has reached an advanced stage.

Indentation tests would show whether parts subject to alternating stresses were in an advanced state of cold work or not.

Overloaded metal members may sometimes be annealed for the removal of the internal stresses at temperatures below those likely to cause change of the original metallic structures. For this purpose the temperature which must not be exceeded in the anneal is the original tempering temperature. Sometimes, as with Duralumin, the tempering temperature was at room temperature or only a little above. A satisfactory anneal would usually disturb the original heat treatment, and the thing to do in such cases is to anneal at the solutioning temperature, completing the heat treatment afterwards.

Steels originally normalized will require re-normalizing.

When the original tempering temperature has been above 300° C., it may be quite feasible to anneal at 300° C. or so for the removal of stresses; but to anneal heat-treatable alloys too near their A_c1 points (or the equivalent points if non-ferrous) is to get the compound coarse and the alloy relatively weak and ductile: full heat treatment is obviously preferable.

Any low temperature anneal might have to be of long duration, and a protective atmosphere of producer or coal gas in the furnace is advisable. Packing the part in a box with cast-iron turnings (for steels), or with sand or lime, assists in protection during annealing.

Heavy Damage

So much damage may have been done, and so limited may the facilities of the engineer be in comparison with those of the manufacturer of the part in question, that the difficulty in recovering the original shape and dimensions may be the factor deciding upon rejection of the part. Whenever it is at all possible to restore shape, then hot work at a medium recrystallizing temperature must be resorted to, and the high temperature required for this may force the necessity for a full heat treatment afterwards, or a normalizing as the case may be. Welding, or building up by welding or by electrodeposition, are obvious possibilities where facilities are available and where experience of the job in hand is sufficient.

Grain Size and Recrystallizing Temperatures

Straightening and cold work removal calls for some knowledge of the recrystallizing temperatures and of the grain size likely to result; the finest possible is always wanted, and this is supplied automatically in the finishing by normalizing or heat treatment for optimum properties. In a few instances, however, it might be of advantage to have the following figures—

Some of the common recrystallizing temperatures are—

For metals of commercial purity

Iron	450° C.
Nickel	500° C.
Copper	200° C.
Aluminium	150° C.
Magnesium	150° C.
Zinc	Room Temperatures
Lead	Below Room Temperatures
Tin	,, ,, ,,

The temperatures given are those below which recrystallization is slow and rather coarse, and above which it is rapid though finer —*when the cold work is medium to heavy in both cases* (grain will not *remain* fine, however, in any case when high temperatures are retained unnecessarily).

At such temperatures it will usually be impossible to remove the cold work entirely. Higher temperatures are called for to start the recrystallization of metal which is but slightly cold worked; higher temperatures are needed when more impurities are present, as these raise the range of recrystallizing temperatures a great deal for small quantities, but not to the same extent for increasing quantities of additional components or impurities.

The recrystallization of strain-damaged *alloys* takes place, therefore, at temperatures somewhat above the temperatures required for the greatly predominating metal. Thus—

Mild steel . Medium carbon	500° C. to 600° C.
Monel metal .	700° C.
Alpha brass .	300° C. to 400° C. (700° C. dead soft)
Beta brass .	450° C. to 550° C.
Elektron, etc. .	300° C. to 350° C.

High carbon steel wire is usually re-softened after hard drawing by what is known as "patenting," which is heating to between 800° C. and 1000° C. and cooling rapidly in air, producing sorbite.

In the case of most of the common heat-treatable non-ferrous alloys the recrystallizing temperatures are above the tempering temperatures, but the softest condition of these is obtained after cold work by the equivalent of spheroidization in the case of steels.

Rapid Deformation by Hot Work may, in general, be carried out without fear of mechanical damage at temperatures from 100 to 200 degrees higher than the lowest above mentioned, and should be finished at the lowest. Magnesium-base alloys must be hot worked in the range 300° to 350°. Duralumin, Hiduminium, etc., at about 470° C.

The Mechanism of Grain Refinement

Upon the analysis of all methods of obtaining fine grain in a metal it would seem that the common features are firstly the provision of *a very large number of centres* of crystallization all at the same time (or within as short a space of time as possible), and secondly the subsequent avoidance of any temperature at which *grain coalescence* (the growth of large grains at the expense of small grains) will take place.

The first part of this programme is carried out at some *change of phase point—or rather beyond it* upon overshooting it so as to encourage a voluminous labile shower of crystallization centres. The second part is served well if, as soon as possible after crystallization from these centres, the temperature is dropped as quickly as possible to inhibit the natural coalescence of the small crystals into the larger ones.

Such a change of phase point may, in general, be reached and exceeded or overshot either by heating or by cooling. Most of the change of grain size in normalizing steels is (for some reason not yet quite clear) upon the heating up change from alpha to gamma iron, and much less of it on the subsequent cooling down from gamma to alpha. Again, a recrystallization of the amorphous-like phase in cold worked metals is upon the heating up, of course, because this change is not reversible. An irreversible change of the opposite kind is that of the liquid to solid transformation on freezing; here it is only the cooling which produces crystallization, and here again the change point must be overshot before the labile shower will give rise to fine crystal grain, .and subsequently the temperature must be kept low to avoid the coalescence which is a perfectly natural phenomenon at the higher temperatures.

In the case of the recrystallizing of cold worked metal, heating to the temperature at which recrystallization will just commence very slowly is rather like the heating exactly to an allotropic change point, or at most not beyond the metastable range around that point; the change will take place slowly but coarse grain results. By rapid heating to higher temperatures than these minima, the change is more rapid from the very numerous centres of the labile shower, hence the grain is fine; but these higher temperatures are to be avoided for a longer period than is strictly necessary for the completion of the change, otherwise any earlier formed slightly larger grains will have time to grow at the expense of the later and smaller ones.

The difficulties presented by the case of a piece of metal cold worked to different extents in different localities cannot be ignored, because they are serious; the temperature high enough for the recrystallization of the least cold worked parts is so much too high in some cases for the more cold worked parts that, unless operations are smartly conducted with a full knowledge of what to expect, crystal coalescence in the latter will have actually preceded the commencement of crystallization in the former. If, however, the temperature for the recrystallization of the less cold worked parts is not reached, nevertheless the amorphous-like metal of these parts will be changed to crystalline metal on its *absorption*, when crystals from other parts grow at its expense into these less cold worked parts; so that, though recrystallization will have taken place, it will be very coarse in the less cold worked parts when their own recrystallization temperatures have not been attained.

Certain additions in extremely small amounts have proved to affect very favourably the volume of the labile shower beyond the change points, or else to inhibit strongly the tendency to coalescence of the original crystals from that shower. At least, the foregoing is one way of expressing the fact that tiny quantities of some elements alloyed with metals or alloys procure fine grain, and there is no accounting for the mechanism of their behaviour except in such terms. Titanium and manganese in the aluminium and magnesium alloys have such effect; somewhat larger quantities of nickel in steels will also give rise to the same grain refining.

A conceivable explanation of such phenomena may be that the liquid solubility of, say, titanium in aluminium is an example of conjugate solutions, or of two phases in equilibrium, in the form of an emulsion of an extremely small quantity of the titanium-rich phase finely dispersed throughout a very large quantity of the aluminium-rich phase. The presence of many millions of these titanium-rich particles might provide centres of crystallization at or beyond the freezing range of the aluminium alloys containing very small quantities of titanium.

GLOSSARY OF METALLURGICAL AND OTHER TERMS AND EXPRESSIONS USED

Age-hardening. Hardening spontaneously on lapse of time.

Age-hardening (artificial). Hastening "age-hardening" by "temper-hardening."

Allotriomorphic. Having external shape not corresponding to the internal structure. (Refers to crystal grains.)

Allotropic Form. A form having a specific atomic arrangement (crystalline or amorphous) to which is due the properties of the form. (The same element may have different properties depending on its allotropic form.)

Alloy. A coherent mixture (or solution) of a metal with one or more other metals, metalloids, or non-metals, often including one or more metallic compounds.

Alloy Steel. A steel containing special elements deliberately alloyed.

Alpha Iron. A body-centred cubic allotropic form of iron.

Aluminium-base. Having aluminium as the chief component.

Amorphous. Structureless; non-crystalline; lacking an orderly arrangement of atoms.

Amorphous-like. Near-amorphous; resembling amorphous.

Anneal. To soften by heating followed by slow cooling.

Anodize. Oxidize the surface electrolytically to resist corrosion.

Apparent Strength. "Nominal strength"; maximum stress of the tensile test as calculated without allowing for the reduction in area of cross-section of the specimen.

Atom. Used in the text to mean: The smallest possible particle of a metal or other element; the unit of a lattice, if the mass is crystalline.

Atomic Weight. The ratio of the mass of an atom of a metal or other element to the mass of an atom of hydrogen.

Austempering. Obtaining superior properties in steels by certain special quenching methods.

Austenite. The solid solution of carbon and/or other metals in gamma iron.

Beta Iron. The non-magnetic form of alpha iron.

Binary. Consisting of two metals (e.g. binary alloy).

Blow-holes. Cavities in cast metals due to air or other gas.

Body-centred. Having a lattice atom (or unit) in the centre of the cube.

Brinell Number. Hardness stress in kilograms per square millimetre.

Brittleness. The opposite of (ordinary) toughness.

Carbide. Compound of carbon and other element or elements; usually iron carbide, cementite.

Carbide of Iron. Straight carbon steel cementite, Fe_3C.

Carbon Steels. Plain steels, without special alloyed elements.

Carbonaceous. Containing carbon in some form.

Carburization. Process of putting extra carbon into steel, particularly into the surface layer.

Case-hardening. Hardening the outside of steel by carburization or by nitriding.

Casting Alloys. Alloys commonly used as cast, without subsequent hot work.

Cathode. The electro-negative pole of an electrolytic cell on which metals may be deposited by electricity.

Cavities. Holes appearing in metals from a variety of causes.

Cementation. The process of increasing the carbon content of (usually) wrought iron.

Cementite. Iron carbide Fe_3C, in a carbon steel; composite carbides in an alloy steel.

Chill Casting. A casting made in a mould which rapidly extracts heat from the metal.

Civil Designation. The name, letters and number given to an alloy by the producing firm.

Cold Work. Hammering, rolling, drawing, drilling, or machining, etc., at temperatures *below* the recrystallizing temperature.

Colloidal. Consisting of tiny aggregates (colloidal particles); not crystallized.

Component. Ingredient; identifiable part of the whole.

Compound (Metallic). A substance (entirely different in all respects from any of its components, formed by the chemical union of the atoms of its components) the smallest particle of which is a molecule, or group of atoms.

Constituent. A component or phase of an alloy identifiable under the microscope; a eutectic is classed as a "constituent," though it consists of two phases.

Constitution Diagram. Temperature-concentration chart of arrest points in the cooling curves of a complete series of alloys of two metals (or of a metal with a compound, or with non-metal) when the cooling is carried out very slowly.

Copper-base. Composed chiefly of copper.

Cored Crystal. Solid solution crystal grain exhibiting lack of homogeneity due to metallic concentrations differing from interior to exterior.

Corrosion Resistance. Capacity to resist corrosive attack.

Creep. Plasticity exhibited on prolonged loading at low stress.

Critical Point. Temperature at which some allotropic change commences, ends, or takes place.

Critical Range. Range of the temperature over which the whole of the allotropic change takes place.

Crystallite. An extremely tiny crystal—much smaller than the common crystals of metals.

Cyaniding. Case-hardening in molten sodium cyanide.

Damping Capacity. The property of damping out stress oscillations.

Decalescence. Absorption of heat as alpha iron undergoes allotropic change to gamma iron.

Decarburizing. Removing carbon from (usually from the surface of) steel.

Dendritic (structure). Resembling a fern growth; cored or dendritic structure; dendritic crystals of metal as cast, etc.

Diamond Hardness. Hardness as determined by the Vicker's machine.

Dispersion. Degree of separation, or degree of fineness.

Ductility. Percentage elongation of a two-inch test piece at fracture; a measure of plastic deformability of all kinds.

Elastic Limit. Stress up to which metal will return at once to its original dimension when the load has been released.

Elastic Modulus. The ratio of stress to strain within the limit of proportionality.

Electrolyte. The liquid of an electrolytic cell.

Elektron. Magnesium-base light alloys having 8 to 10 per cent aluminium, 1 to 5 per cent zinc, and about 0·5 per cent manganese; a proprietary name or trade mark for a series of these alloys.

Emulsion. A uniform mixture of two phases, e.g. a mixture of a solution of B in A and a solution of A in B.

Endurance Range. The algebraic difference between the upper and the lower limits of stress applied in a fatigue test.

Equiaxed. Having all three axes equal in length; approximately cubic or rounded, like grains.

Equilibrium Diagram. See Constitution Diagram.

Etching. Eating into the surface by reagents to distinguish microscopic constituents.

Eutectic. The last freezing constituent; the most fusible alloy; a two-phase type of constituent derived from the liquid to solid change of state.

Eutectoid. A two-phase type of constituent like a eutectic but derived from solid solution.

Extensometer. An instrument for measuring extremely small tensile strains.

Face-centred. (Of a cubic lattice.) Having an atom (or unit) at the centre of each face of the cube.

Fatigue Resistance. That maximum stress which, when applied as reversals at speed, only just fails to cause the fracture of the metal in ten million reversals.

Ferrite. Either free or eutectoid alpha iron (the alpha iron of martensite, troostite and sorbite is not described as ferrite).

Ferrous. Composed chiefly of iron.

Flocculate. To form loose aggregates, or clusters.

Flow Structure. The appearance of directionality in a mass (as of flow) following plastic deformation.

Frit. To bond together by causing incipient fusion of the surfaces of the particles of a compressed mass (compare Sinter).

Gamma Iron. Iron in the gamma allotropic state, face-centred cubic.

Ghost Lines. Pale flow-structure lines in rolled steel resulting from the extension of areas deficient in pearlite (due to the presence of phosphorus).

Gliding Planes. Preferential planes of shear in the crystal lattice.

Grain. A particle or mass of metal having but one lattice orientation only; an irregular polygonal mass of metal of size and shape which depends on the accidental interference of crystal growth from many centres in a metallic mass at the same time.

Graphitic Carbon. Carbon in the allotropic form known as "graphite."

Graphitization. The change of cementite to graphite.

Hardening. Giving hardness (also strength and high elastic limit) usually by quenching, as of a steel.

Hardness. A mechanical property of materials; a special stress property of resisting the deformation of penetration and wear.

Heterogeneous. Consisting of (mixed but undissolved) different components.

Hiduminium. "High Duty" alloys of aluminium; a proprietary name or trade mark of certain aluminium alloys.

High Duty Alloy. See Hiduminium.

Homogeneous. Consisting of one uniform component or single solution of components.

Hot working. Extruding, stamping, pressing, forging, rolling, drawing, etc., at temperatures above the recrystallizing temperatures.

Hyper- Above, e.g. hyper-eutectoid, above 0·9 per cent carbon.

Hypo- Below, e.g. hypo-eutectoid; below the eutectoid composition; below 0·9 per cent carbon.

Hysteresis. A lagging or hanging behind; e.g. the differences in temperature between the cooling change points and the corresponding heating change points.

Impact Resistance. The number of foot-pounds to fracture a specified test-piece in the Izod or other similar "Impact" testing machine.

Inclusions. Unwanted, accidental, or unavoidable slag, sulphide, etc., particles or gas holes in a metal.

Inhibit. To prevent; to check.

Inverse Rate Curve. A chart or graph of the temperature, and of the time taken to cool through a stated number of degrees at that temperature.

Irreversible Steels. Steels which, owing to large hysteresis, exhibit hundreds of degrees difference between A_r points and the corresponding A_c points.

Labile. Of a structural state corresponding to the state of unstable equilibrium in mechanics; of a state capable of changing even after the causative change influence has ceased to exist.

Lamellar. Consisting of layers of different components.

Laminated. Consisting of layers.

Lattice. The atomic structure or the internal architecture of crystalline metal.

Limit of Proportionality. The stress up to which the stress/strain ratio is constant.

Liquidus. The equilibrium diagram line above which the alloys are in the liquid (molten) state.

Lueder's Lines. Lines appearing on the outside of a tensile test specimen (which has been plastically distorted) indicating that shear is taking place in tension.

Macrosection. A section cut, polished, and etched to exhibit coarse metallic structure such as "flow" or "ingot-piping."

Martensite. The first "transition product"; a molecular dispersion of iron carbide in alpha iron.

Matrix. The ground-mass, or last-freezing constituent.

Metalloid. An element having properties between a true metal and a non-metal.

Metallurgy. The technology of metals.

Metastable. Of a structural state corresponding to the state of neutral equilibrium in mechanics; of a state capable of changing only so long as changing influence is exerted.

Microsection. A section cut, polished, and etched to exhibit microscopic structure.

Molecular. Consisting of molecules, or groups of combined atoms.

Nitriding. Case-hardening by ammonia gas.

Nitrogen Atmosphere. No oxygen present, only inert nitrogen; a non-oxidizing atmosphere.

Nitrogen Hardening. See Nitriding.

Nominal Strength. See Apparent Strength.

Non-metal. An element with properties very different to those of a metal; e.g. sulphur.

Non-reactive. Unable to combine chemically with some other, or any other, substance.

Normalizing. Heat treating (i) steel—to produce fine grain and to relieve internal stresses, (ii) Duralumin, to solution the compounds and to quench afterwards.

Notch Sensitiveness. The opposite of notch toughness.

Notch Toughness. The property of resisting the propagation of a crack once formed; conveniently measured in an Izod type machine with a fine notch in the specimen.

Nucleus. A centre; a tiniest particle; a beginning.

Occluded. Condensed upon a surface, or dissolved in the surface.

Optimum Properties. Best strength hardness, etc., for any given degree of ductility and notch toughness, or vice versa.

Ore. The impure metal, or the oxide, sulphide, or other compound of the metal, which is mined from the earth's crust (together with earthy impurities) as a raw material for the preparation of the pure metal.

Oriented or Orientated. Turned in one direction; facing one way.

Oxidize. Cause to combine chemically with oxygen.

Passive State. Reactivity checked by filming over with some product of early reaction, e.g. by an oxide film formed in nitric acid.

Pearlite. A constituent of steels; a eutectoid consisting of lamellar cementite and ferrite.

Pegging. Preventing or checking slip between two surfaces by the insertion of "pegs" or keys registering with both surfaces.

Permanent Set. An early increment of plastic strain produced as a result of exceeding the elastic limit of stress.

Plastic Bronze. Bronze-lead solid emulsion.

Plastic Deformation. Change of shape beyond the elastic range.

Polycrystalline. Containing many crystal grains.

Porosity. A condition of extremely fine holes or pores.

Precipitate. A substance thrown out of solution; caused suddenly to form (of a constituent).

Precipitation. The act of throwing out of solution, or of forming from a solution.

Primary. The constituent which first forms on freezing: prior to the eutectic formation.

Pro-eutectoid. Prior to eutectoid formation: not of the eutectoid itself.

Proof Stress. A minor mechanical property: the stress causing 0·1 per cent strain (permanent) when induced for 15 seconds.

Pyramid Hardness. Hardness by the Vicker's Testing Machine. (See Diamond Hardness.)

Pyrometer. Thermometer for measuring the higher temperatures by optical or electrical means.

Quenching. Cooling at definite rates (usually fast) by plunging whilst hot into different media.

Real Strength. Maximum stress actual: maximum load per unit of actual area; stress calculated by making allowance for the reduction in area of the tensile test specimen.

Recalescence. The evolution of heat when gamma iron undergoes allotropic change of state to alpha iron.

Reduce. To take oxygen away from a compound; to convert an oxide of a metal into a metal.

Refine. To make the metallic structure, particularly the grain size, finer.

Season Cracking. The breaking up of severely cold-worked metal owing to the combined effects of corrosion and internal stresses left after cold working.

Secondary. Belonging to the eutectic.

Segregation. Collecting together around a centre; massing together.

Shock Resistance. See Impact Resistance.

Sinter. To produce incipient fusion in a dusty or powdery mass so that it is consolidated.

Slag. Glassy product formed in melting or smelting.

Slip Bands. Visible marks of slip between blocks of crystalline metal.

Softness. The opposite of hardness.

Solid Solution. The finest possible (atomic) mixture of two or more solid substances; atomic mixture in the solid state; one crystalline lattice contains all the components as atoms.

Solidus. The line of the equilibrium diagram below which the alloys are solid.

Solute. The substance dissolved in the solvent.

Solutioning. Heating up to, and holding at, the temperature required to break up and dissolve (in solid solution) any compounds of the metals which have been precipitated by previous treatment.

Solvent. The medium dissolving the solute.

Sorbite. The third "transition product" of steel, containing the finest crystalline dispersion of the iron carbide.

Space Lattice. See Lattice.

Spheroidization. Heating and holding just below the A_c1 point to permit segregation or "balling up" of the cementite laminae of the pearlite.

Stellite. A cast weldable heat-resisting alloy of cobalt, chromium, tungsten and carbon.

Straight Carbon Steel. See Carbon Steel.

Strain Ageing. The hardening and strengthening effect of strain which increases with lapse of time following straining.

Strain Hardening. Hardening and strengthening due to the presence of "amorphous" metal of cold work when plastic strain has taken place.

Strength. A mechanical property commonly measured by the apparent or nominal strength, or "ultimate tensile stress."

Stress-Strain Diagram. A chart or graph showing the relationship between stress and strain as loading is increased from zero up to the point of fracture.

Structure. The arrangement or disposition of the constituents, components, or atoms of a mass.

Sub-metal. See Metalloid.

Sub-microscopic. Too small for the microscope to resolve.

Superlattice. A new lattice imposed on an original lattice by foreign atoms (in solid solution) taking up *regular* positions.

Supersaturated. On the point of precipitation; overdue for precipitation.

Temper-brittleness. The notch sensitiveness sometimes arising from tempering certain steels at low tempering temperatures.

Temper Hardening. Raising the temperature to induce a compound precipitation which hardens the metal.

Tempering. Raising the temperature to soften a quench-hardened metal.

Toughness, Notch. See Notch Toughness.

Toughness, Ordinary. The strength-ductility product.

Transition Product. A stage of the change from one equilibrium state to another.

Troostite. The second "transition product" in steel.

Undercooled. Cooled below the equilibrium temperature of change without the change having taken place; overdue for some new phase to form on cooling.

Units of Lattice. The atoms—in some cases groups of atoms—of which the lattice is constituted.

Valency. The habitual chemical combining capacity of an element, which may be expressed in terms of the number of chlorine atoms with which one atom of the element habitually combines. Thus, one atom of aluminium combines with three atoms of chlorine in the usual way—not with one, two or four atoms; aluminium is, therefore, called "trivalent," or is said to have a valency of three. All metals and other elements tend to combine chemically according to such valencies.

Viscosity. The resistance offered by a liquid to flow or to shear.

V.P.N. Vicker's Pyramid Hardness Numbers.

Weakness. The opposite of strength.

Widmanstätten. A distinctive visible structure produced within a crystal grain by the development of some new phase on the octahedral planes of the cubic lattice.

Wrought Alloys. Alloys commonly used not in the cast condition, but only after hot working.

Yield Point. A minor mechanical property of metals; the stress at which strain increases rapidly without further increments of stress.

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(See also Glossary of Metallurgical Terms, page 143.)