Chapter II

FUNDAMENTAL CHARACTERISTICS OF STEELS

The element, iron, when its impurities are reduced to a few hundredths per cent, is only slightly harder and stronger than copper. Without too great effort such iron is producible in a commercial way but the tonnage represented by such material is relatively small. Its exceeding plasticity and softness are for many applications a disadvantage rather than an advantage and accordingly the vast amount of iron as known commercially is a moderately impure metal, and the processes of winning the iron from the ore and refining it to steel are not designed primarily with the delivery of a highly purified element as the main objective. In the interest of high yield from the ore the reducing agent, carbon, is employed plentifully, and later the carbon is eliminated to the desired extent by oxidation, along with such other elements as were originally reduced from the ore and are oxidizable. It is this same inexpensive element, carbon, seemingly, which influences the properties of iron more profoundly than any other and often more usefully. It is *allotropy* in iron, however, and its attendant circumstances, particularly with reference to carbon solubility, which makes possible the unparalleled range of properties which are encompassed by steels covering the usual range of carbon content; for allotropy is retained in the presence of considerable carbon and other elements, and thereby the capacity for advantageous heat-treatment comes into play.

Allotropy and Binary Alloys of Iron — The metal iron, as shown in Fig. 1, exists in two isometric crystal forms (1) alpha (*a*) and delta (δ) iron whose solid-solutions are called ferrite (or delta ferrite) and (2) gamma (γ) iron whose solid-solution is austenite. The alpha form of pure iron exists below about 1670 degrees Fahr. (910 degrees Cent.) and above about 2552 degrees Fahr. (1400 degrees Cent.) when it
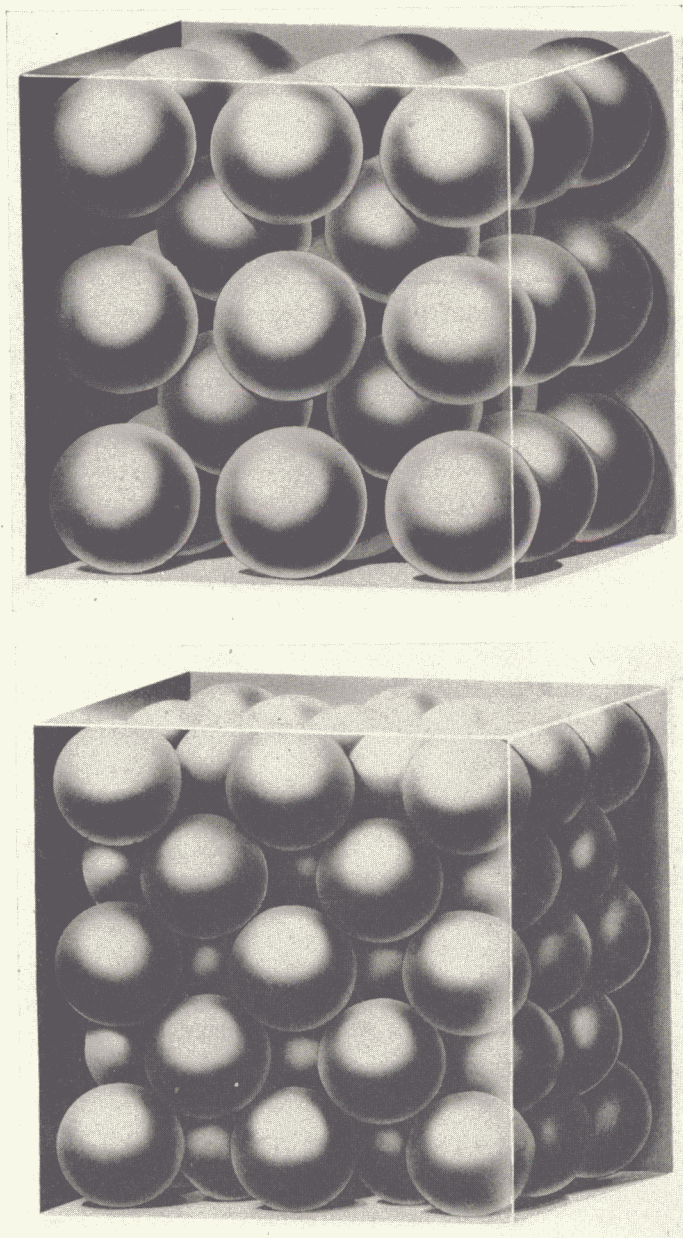
Fig. 1—The Arrangement of Atoms in the Two Allotropic Forms of Iron; Above, Alpha (Delta) Iron; Below, Gamma Iron. *(Courtesy R. M. Bozorth, Bell Telephone Laboratories).*

is called delta iron. Gamma iron exists at the temperatures between these two ranges. The various alloying elements have widely different solubilities in each of these two forms of iron, a circumstance which brings about unique temperature ranges for the transformations in the presence of added elements. In pure iron the change from alpha iron to gamma iron occurs at a single temperature as is the case, for example, when a pure substance melts; the presence of any additional elements creates a more or less narrow temperature range over which both forms of iron may exist simultaneously in equilibrium. This circumstance, indicating a partition coefficient other than unity, and the fact that this coefficient changes with the concentration of the alloying element gives rise to characteristic types of changes in the transformation temperatures of iron alloys; there are two primary types with subdivisions in these types. All this is concisely stated in the conventional equilibrium diagram, and for reference the types are illustrated here in Fig. 2. It is a matter of grave doubt that these binary equilibrium diagrams are of any considerable practical interest to the student of alloy steels for two reasons:

1. The steels of commerce largely contain carbon and other elements which greatly alter the temperatures and compositions involved, except in the case of certain very low-carbon alloys with especially high-alloy content.

2. The majority of steels, as heated and cooled, actually transform at temperatures far removed from equilibrium temperatures because of their very slow rates of reaction near these temperatures. Furthermore, the resulting structures are not predictable from any equilibrium diagram, and as will become clear, the really valuable knowledge concerning alloy systems is that relating to structure.
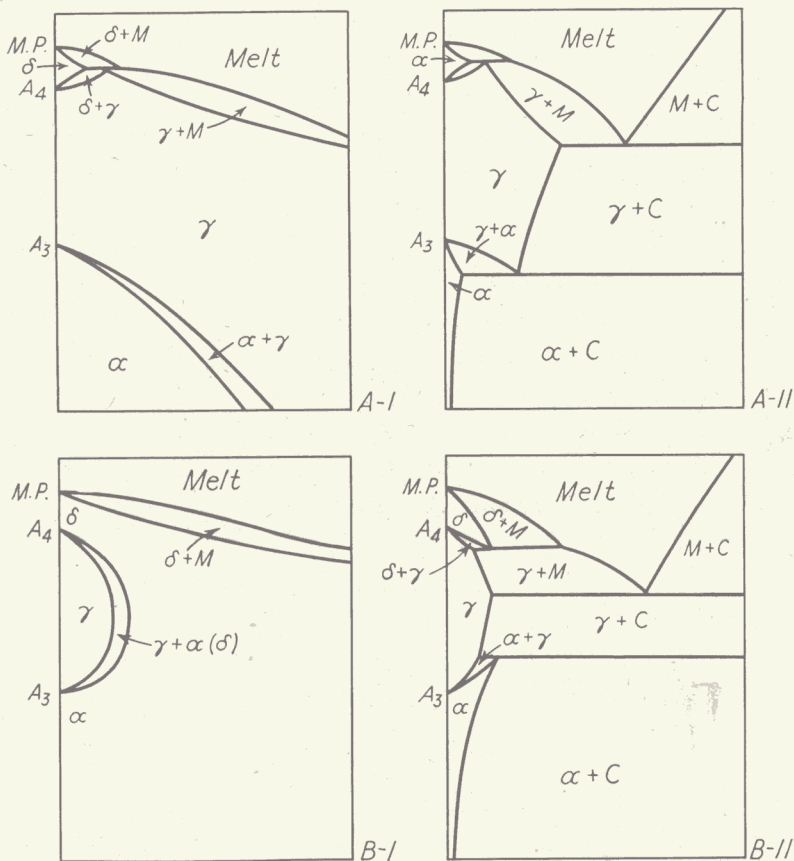
Fig. 2—The Two Types, A and B, and the Subdivisions I and II of Phase Equilibrium Diagrams for Iron Alloys. *(After Wever).*

The characteristics of the transformation in the various binary alloys permit a classification as in Fig. 2, as follows:

Type A — Division I. The alloying element widens the temperature range for stable austenite by depressing the alpha-gamma transformation and raising the gamma-delta transformation temperature. The system shows no iron-rich compounds (or solid-solutions in the alloying element) at concentrations encroaching upon the alpha or delta fields. The two-

phase ($\gamma + \delta$) region reaches the melting range and the two-phase ($\alpha + \gamma$) zone is depressed toward ordinary temperatures. Examples: Mn, Ni and Co.

Type A — Division II. Same as Division I above except iron-rich compounds (or the solid-solution in the alloying element) become stable at compositions encroaching upon the alpha or delta phases. Examples: Cu, Zn, Au, N and C.

Type B — Division I. The alloying element narrows the temperature range for stable austenite and finally renders it non-existent. The austenite zone is completely surrounded by a two-phase ($\alpha + \gamma$ or $\gamma + \delta$) field, that is uninterrupted by intermetallic compounds or solid-solutions in the alloying element. Examples: Si, Cr, W, Mo, P, V, Ti, Be, Sn, Sb, As and Al.

Type B — Division II. The same as Division I above except that intermetallic compounds or constituents other than the alpha iron and gamma iron solid-solutions make their appearance interrupting the "loop" with its enveloping two-phase zone. Examples: Tantalum, zirconium, boron, sulphur and cerium.

There is one, and there may possibly be two, interesting special instances although they are listed as regular cases above. Chromium up to some 7 or 8 per cent lowers the temperature range of the alpha-gamma transformation; further increments then raise it. However, the lowering of the gamma-delta transformation temperature in the range up to 8 per cent chromium is more rapid than that of the alpha-gamma temperature and it is consistent therefore to regard that chromium uniformly narrows the austenite range.

Cobalt probably exemplifies the opposite case in that it widens the austenite temperature range even though it actually appears to raise the alpha-gamma transformation temperature with small concentrations.
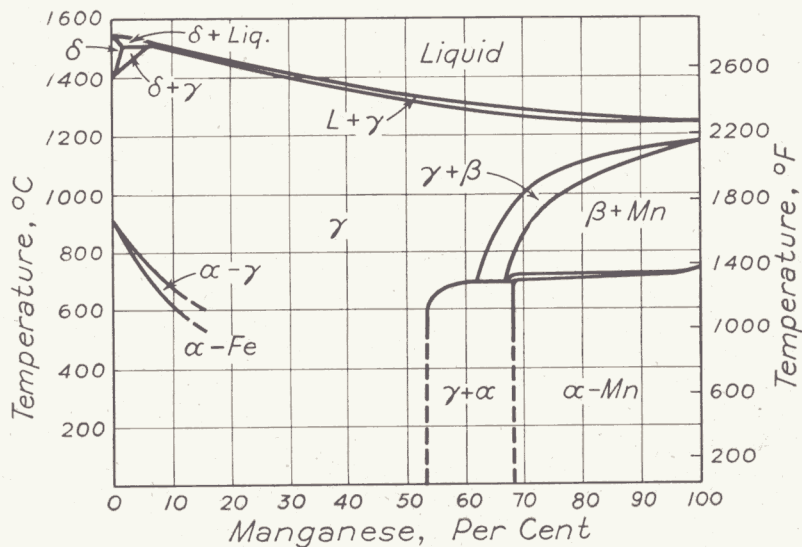
Fig. 3—Equilibrium Diagram for Iron-Manganese Binary Alloys. *(Metals Handbook—1939)*.
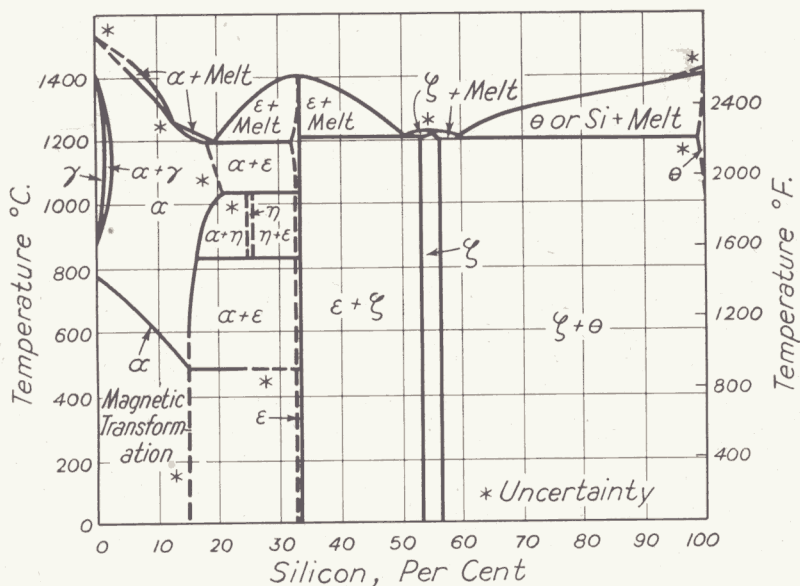

Fig. 4—Equilibrium Diagram for Iron-Silicon Binary Alloys. *(Metals Handbook—1939)*.

Fig. 5—Equilibrium Diagram for Iron-Nickel Binary Alloys.
(Metals Handbook—1939).

The binary equilibrium diagrams for eight common alloying elements with iron are presented in Figs. 3 to 9 inclusive.

Significance of Equilibrium Diagrams — It must be stressed here that with respect to alloy systems, as thus far discussed, the state of affairs at supposed equilibrium has been described. This implies not only that every known means
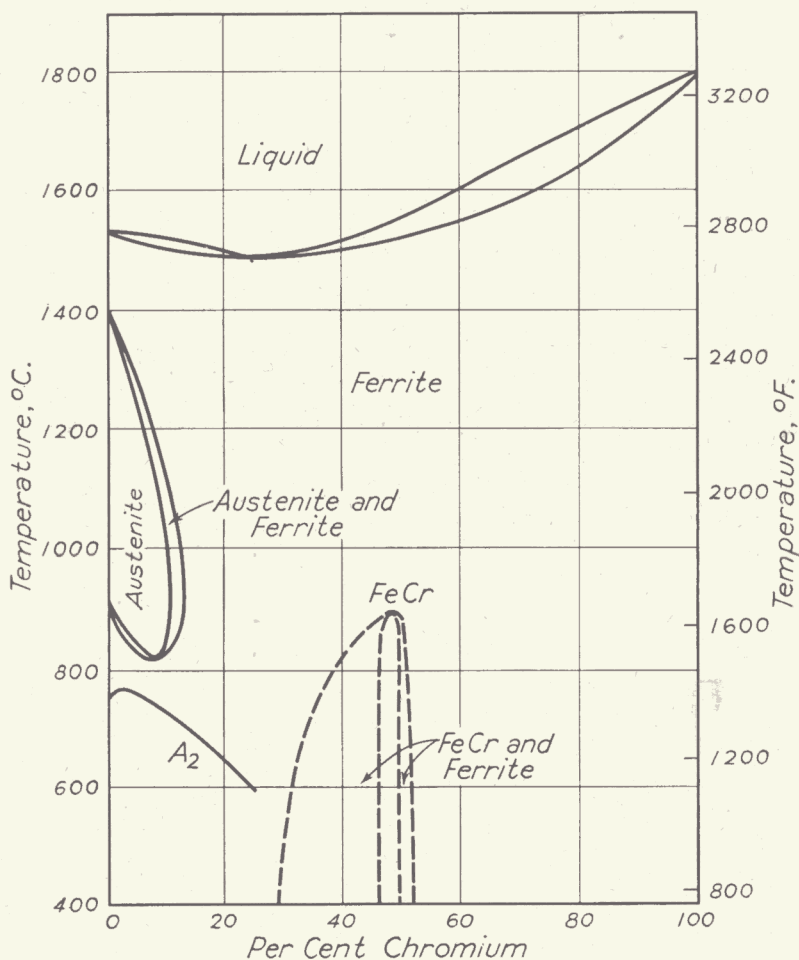
Fig. 6—Equilibrium Diagram for Iron-Chromium Binary Alloys. *(Metals Handbook—1939).*

has been employed of inducing all tendencies toward change from one state to another to assert themselves (save perhaps that of holding the metals at constant temperature for decades or centuries!) but that all such tendencies have actually been fully realized. In many cases the difference in temperature required for a reversal of reactions has been nar-
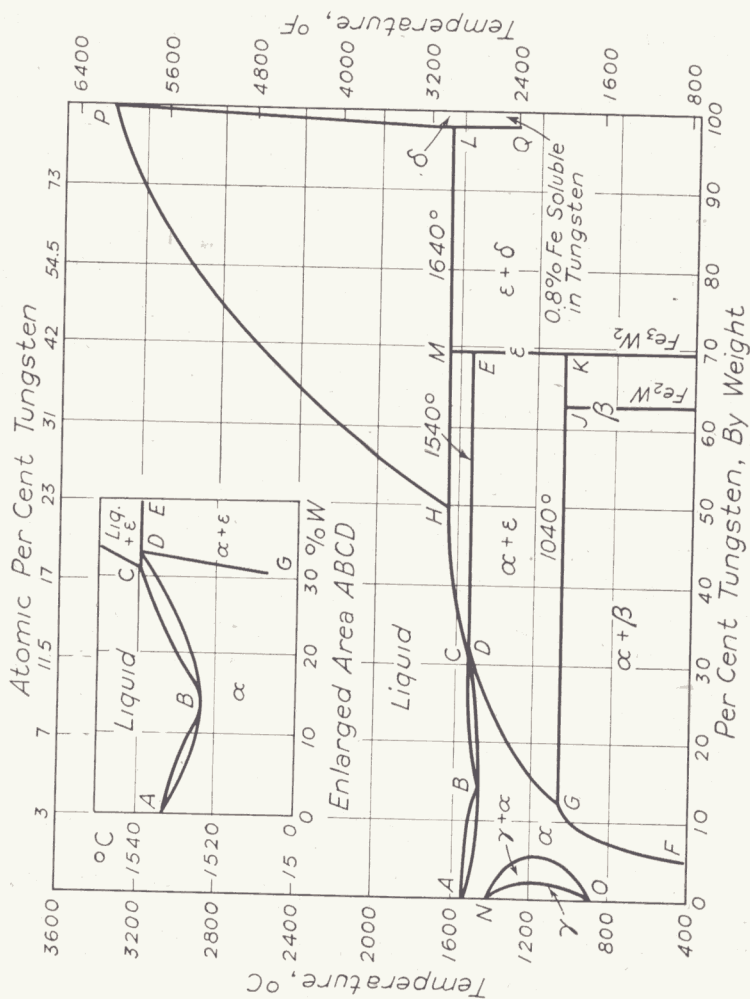
Fig. 7—Equilibrium Diagram for Iron-Tungsten Binary Alloys. (*Metals Handbook—1939*).

rowed down to such an extent that no great error can exist between the true equilibrium and the diagram. But as the enquiry into alloy steels proceeds it will be only too apparent how little the student is concerned with equilibrium. Rather than a concern for what tends to happen, or even ultimately

Fig. 8—Equilibrium Diagram for Iron-Molybdenum Binary Alloys. *(Metals Handbook—1939)*.
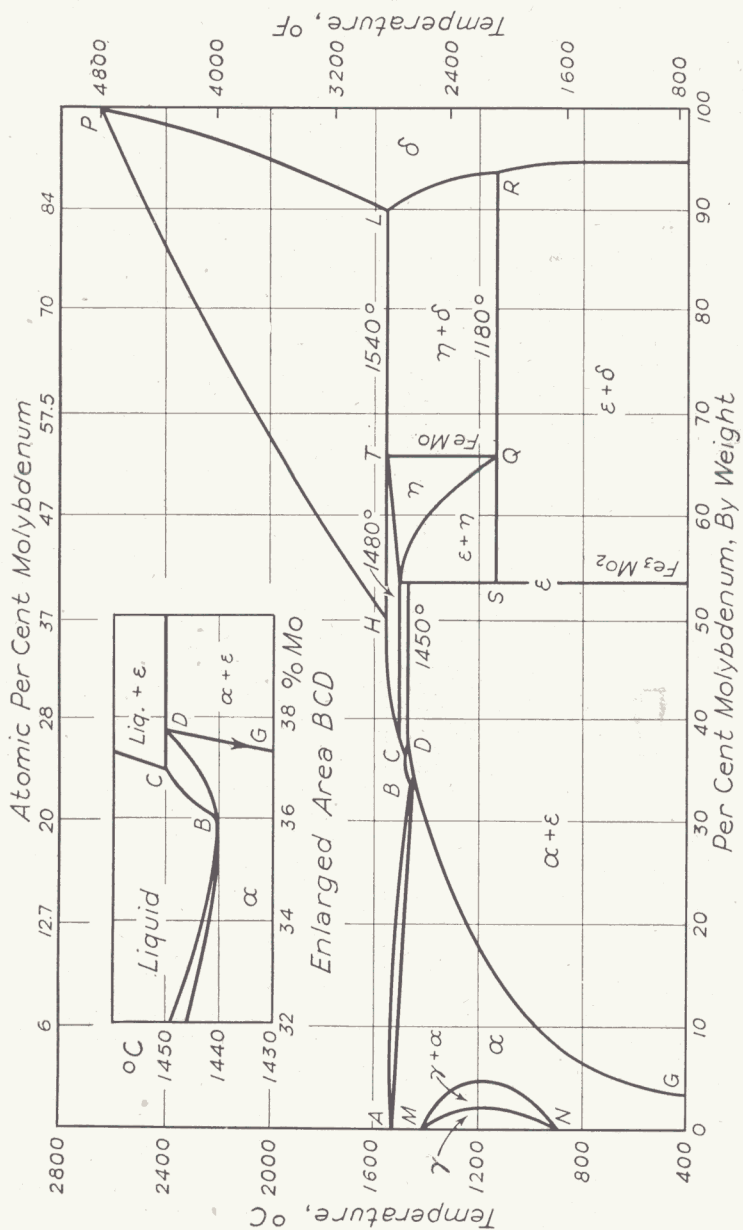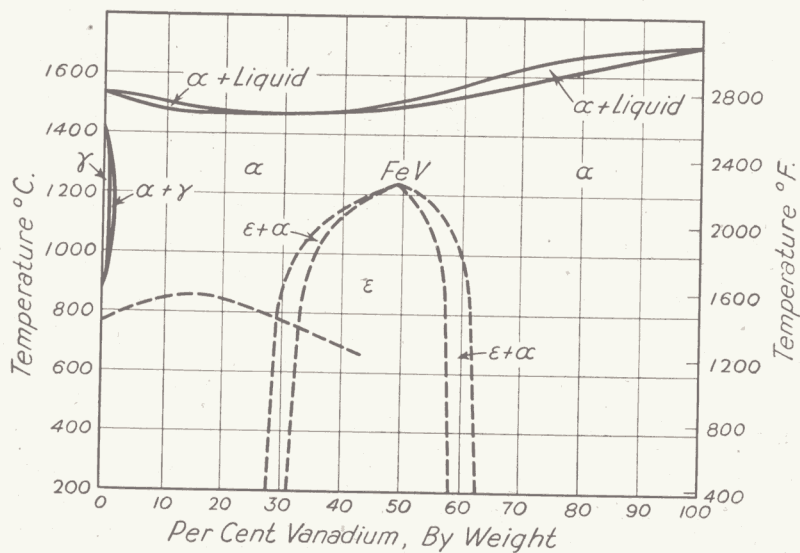
Fig. 9—Equilibrium Diagram for Iron-Vanadium Binary Alloys. (*Metals Handbook—1939*).
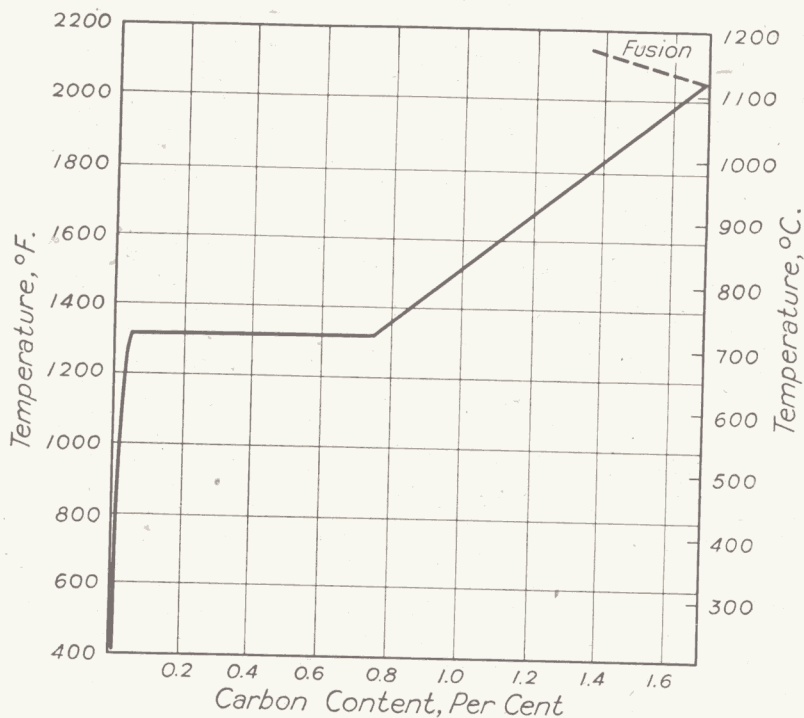

Fig. 10—The Solubility of Carbon in Iron as Influenced by Temperature. Composition is that of Commercial Steel rather than Pure Iron.

would happen, his interest is in what does *actually* occur in a given practical period of time at a certain temperature. Nevertheless a background of knowledge relating to equilibrium in alloy systems is of value in that it enables one to predict that certain changes can never occur in given compositions at given temperatures, even though such knowledge helps not at all in predicting how rapidly, or to what extent, a *possible* reaction will occur.

THE CARBON STEELS

Of the several iron-alloy systems already mentioned, one, the iron-carbon system, is of outstanding importance because carbon is ubiquitous and because relatively large proportions of beneficial alloying elements may be present along with the carbon without much altering the nature of the structures which may be developed or the general pattern of behavior, however much the conditions are altered for producing any certain structure. The so-called plain carbon steels must serve as the prototype of microscopic structures as well as the basis of comparison in evaluating alloy effects, and at the risk of some tediousness it will be profitable to review their characteristics.

The Ferrite-Carbide Aggregate—Carbon steel as purchased to be formed and shaped, and as it usually enters service, is really a mixture of a moderately pure iron with iron carbide Fe_3C (traces of other elements) or cementite. There may be also small particles of nonmetallic substances since steel must be melted and refined in refractories or under slags, the solubilities of which, small as they are, are greater in steel at high temperatures than at low. Indeed it now appears that if there were available a source of chemically pure iron and carbon alloy it would rarely be employed until an addition, a well-advised, judicious addition, of foreign material had been made. Nevertheless, carbon steel of com-

merce is, from the metallurgical standpoint, a dispersion of iron carbide in ferrite. The proportion of iron carbide depends very closely upon the carbon content, for only about 0.03 to 0.04 per cent carbon is soluble in ferrite even at the maximum (around 1320 degrees Fahr. — 715 degrees Cent.), and perhaps at the lowest temperature at which carbon may be rejected from ferrite the solubility scarcely exceeds 0.01 per cent. As a rule, the per cent of cementite in an annealed carbon steel is the per cent carbon content (less 0.02 per cent) times 15. Or one may merely subtract 0.3 per cent from the product of the carbon content times 15.

The Solid-Solution Austenite—Upon heating, the carbon solubility increases, and abruptly, at the temperature of the formation of austenite in the presence of carbon. The solubility of carbon in average carbon steels is shown in Fig. 10. The *rate* at which the carbon dissolves has nothing to do with this diagram which implies a close approach to equilibrium, a state which even in carbon steels is often not fully reached in a matter of hours. To secure a complete conversion of the metallic constituents of the carbon steels to the single constituent austenite requires a heating into the temperature range indicated as shaded in Fig. 11. The very minimum temperature range for the solution of the carbide is seldom employed because the time involved may be rather too long. It is in this wholly-austenitic range that much of the forging and rolling of steels is done, but the higher carbon steels (above about 0.90 per cent carbon) are not necessarily heated in this range for heat-treatment for it is often desirable not to have all the iron carbide dissolved.

The Transformation of Austenite is Controllable — The science of the heat-treatment of steel properly begins with the steel at an elevated temperature with all or most of it in the austenitic condition. Here the austenite has two characteristics of interest: (1) composition and homogeneity of com-
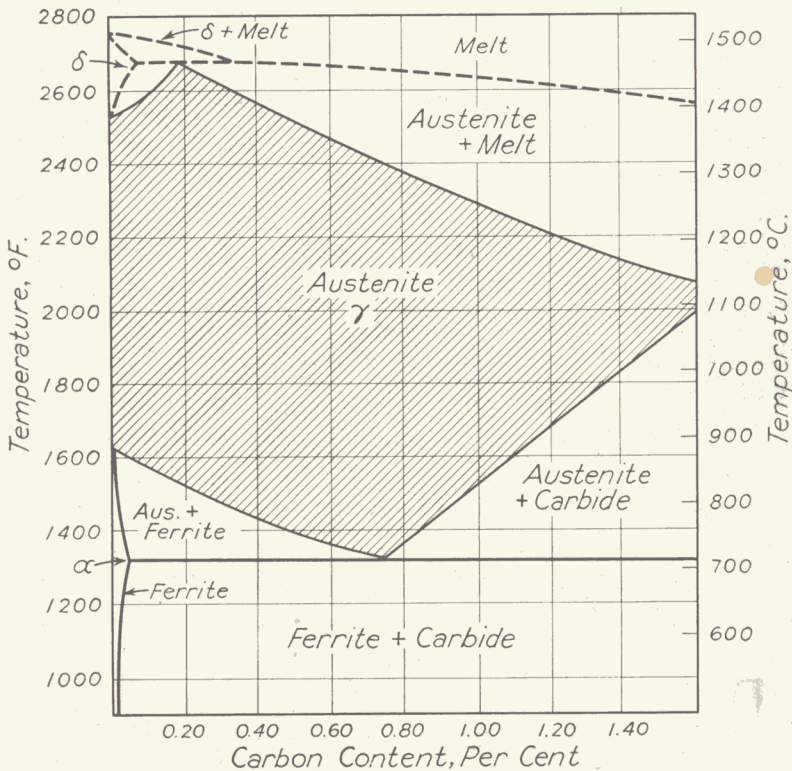
Fig. 11—Temperature and Composition Limits for the Forma-
tion of Pure Austenite (Shaded). Regions for Ferrite and Hetero-
geneous Constitution are also shown. The Diagram is representative
of Commercial Compositions of Carbon Steel.

position and (2) a certain grain size. This austenite will
revert to ferrite and carbide when it is cooled and by adjust-
ing the conditions of cooling, specifically the rate, a control
of the microscopic structure is at hand; that is, the mode of
distribution of the hard carbide constituent in the otherwise
soft ferrite is under considerable control at will. Now one
of the various rates at which steel may be cooled from the
austenitic state is the more or less automatically-determined
rate at which a certain rolled product will cool as laid out on
the cooling bed after the last pass or hot operation. Tech-

nically this is not a heat-treatment, but metallurgically it is, for it is a well reproducible handling for any given section and results in a definite mode of transformation with resultant property acquisition suited to certain uses. Accordingly, the so-called "as rolled" condition must be regarded as resulting from one of the heat-treatments even though no particular effort is required to achieve these rates of cooling.

When carbon steels are cooled at an exceedingly slow rate (a few degrees per hour), the transformation of austenite will occur fairly nearly in accord with Fig. 11, that is, within, say, 5 or at most 10 degrees of the equilibrium temperature, but when cooled at more practical rates the reactions are quite different as to temperature of occurrence. The corresponding equilibrium diagram for the pure iron and carbon system is shown in Fig. 12. The lag in the transformations, — both upon cooling and heating, — in the case of the pure iron-carbon alloys, has been measured by Mehl and Wells[2] as shown in Fig. 13. Equally influenced by rate of cooling is the resulting structure and in this we are intensely interested.

The decomposition of the solid solution austenite does not begin instantly when its temperature is lowered to that at which, in time, it would transform. Instead, there is a definite period of lag which is presumably occupied by nucleus formation or the chance association of sufficient atoms of the new constituent to form a permanent crystallite. At any rate this reluctance is very definite and constant for any particular austenite, and a degree of undercooling is possible which, quite necessarily, depends upon the rate of the cooling.

Modes of Carbide Dispersion — When a new constituent develops within a metal it generally has a wide range as to possible final size of the individual precipitated particles,

---

[2]"Constitution of High Purity Iron-Carbon Alloys", by Robert F. Mehl and Cyril Wells, 1937, Technical Publication, American Institute of Mining and Metallurgical Engineers, No. 798, Iron and Steel Division, No. 175. *Metals Technology*, June 1937.
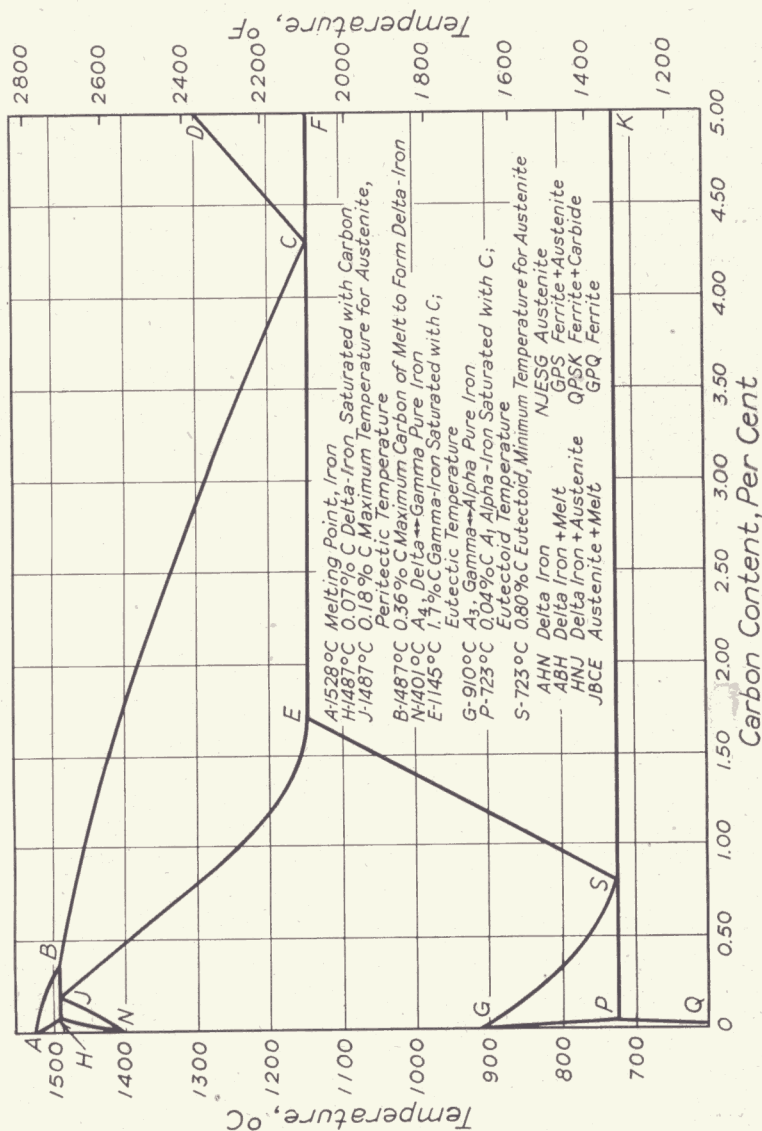
Fig. 12—The Iron-Iron Carbide Diagram.

depending upon the specific diffusivities involved; but in con-
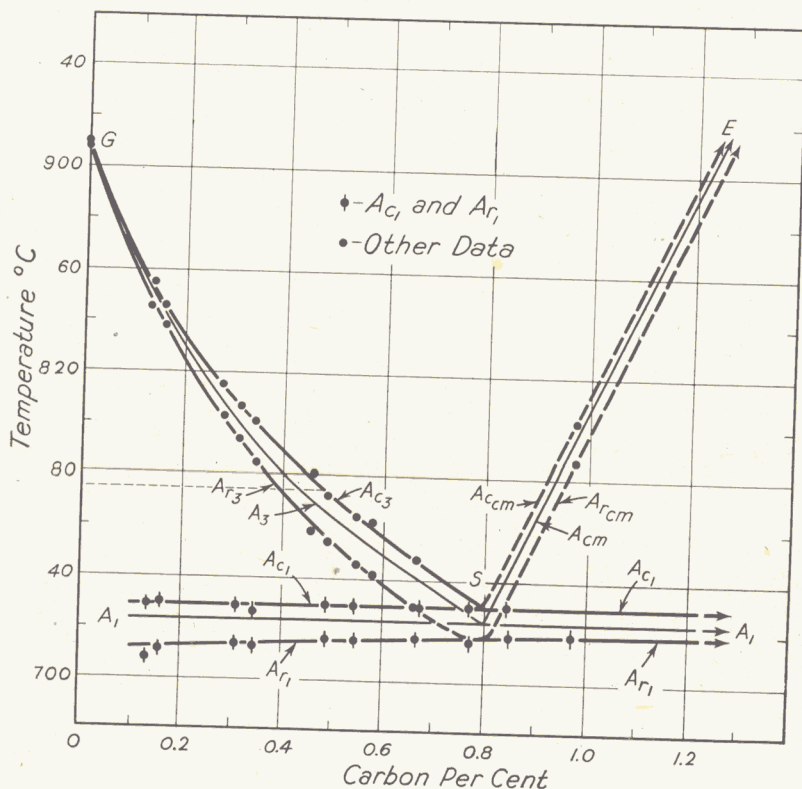tour it generally conforms to two of the three primary cate-

Fig. 13—The Transformation Temperatures in the Pure Iron-Carbon Alloys as Influenced by Heating and Cooling at ⅛ Degree Cent. per Minute. The Probable Equilibrium Temperatures for the Several Phases are also shown. *(Mehl and Wells²).*

gories, — films, filaments or particles. In the case of iron carbide, depending upon circumstances, it may form in plates or lamellae, intergranular cells or somewhat spherical particles, all over a vast range of sizes. Nowhere in metallurgy is there a more complete representation of the effect of the distribution of hard particles in a soft plastic matrix than in steel. The finer the dispersion, the greater the hardness of the aggregate. Thus the extraordinary range of properties is easily accounted for. In the clusters of lamellae the films of hard carbide may be as thin as half of one millionth inch
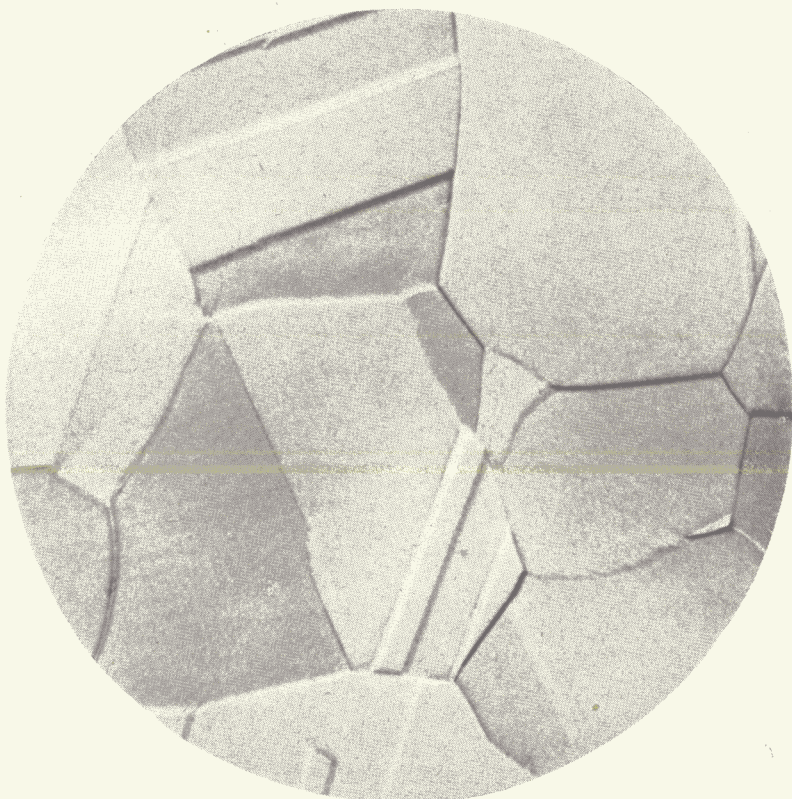
Fig. 14—Microscopic Appearance of the Solid-Solution Austenite. Note Twinned Grains; (1000×). *(Vilella).*

or perhaps as thick as 0.00008 inch. In the case of spheroidal particle size the range is from what might be called molecular size to spheroids perhaps nearly 0.001 inch in diameter, and this arrangement of the disperse system is under control through the cooling and reheating of the steel, — in short, heat-treatment.

The Lamellar Family of Structures —Consider for simplicity a steel containing some 0.70 to 0.90 per cent carbon. Heated to about 1400 degrees Fahr. (760 degrees Cent.) for example, the piece consists wholly of austenite in
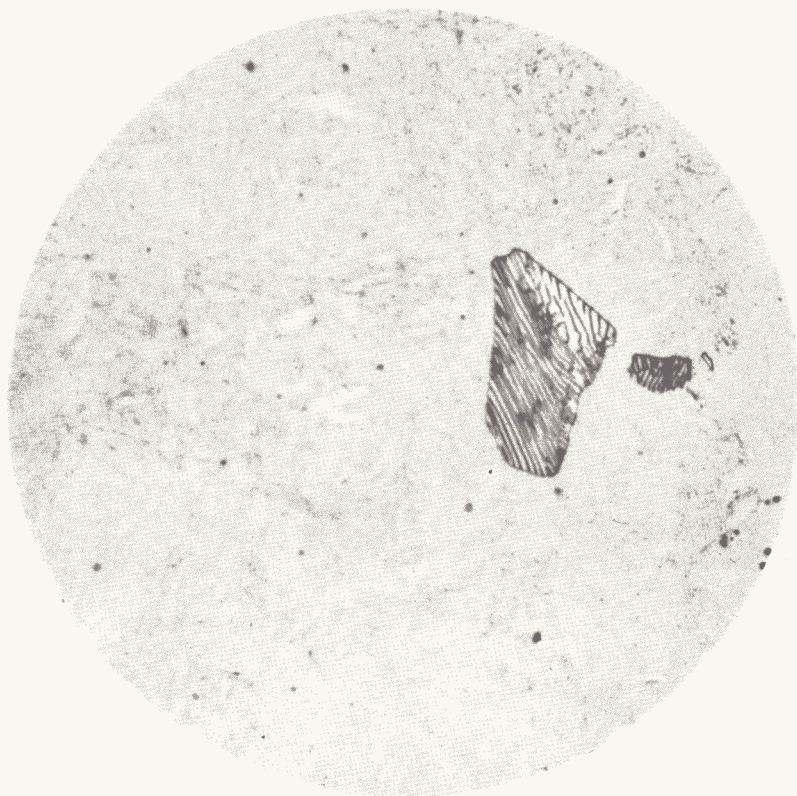
Fig. 15—The Transformation to Pearlite in a Eutectoid Steel Arrested at Its Beginning. Note Nodular Growth from Nuclei and Simultaneous Formation of Both Ferrite and Carbide; (1000×). *(Vilella)*.

polyhedral grains, as in Fig. 14. Upon slow cooling, say at a rate of 4 or 5 degrees Fahr. (2.5 degrees Cent.) per minute, the austenite reverts to ferrite and carbide with the lamellar arrangement. The undercooling will amount to some 15 to 20 degrees Fahr. and the transformation begins at 1300-1305 degrees Fahr. The temperature of the piece may actually rise slightly during the transformation, while the final reaction may even occur at a lower temperature. The lamellae are spaced about 0.000015 inch apart and are about 0.000002 inch

Fig. 16—The Transformation to Pearlite Arrested when 25 Per Cent of the Austenite had Transformed; (1000×). *(Vilella)*.

thick. The hardness is about 210 B.H.N. The mode of transformation may be seen in the series of photomicrographs, Figs. 15 to 19 inclusive.

The effect of increasing the rate of cooling a 0.75 per cent carbon steel may be seen in the following table:

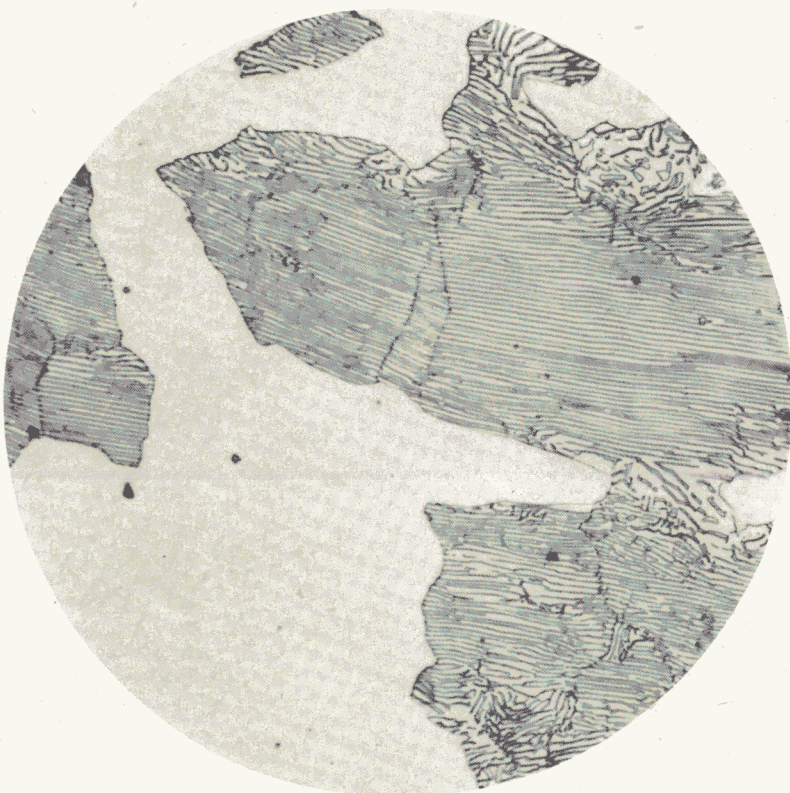| Approximate Cooling Rate: 1400° F. to Transformation	Temperature of Transformation	Brinell Hardness Number
5° per minute	1305° F.	210
230° per minute	1240° F.	315
1800° per minute	1010° F.	415

Fig. 17—The Transformation to Pearlite Arrested when 50 Per Cent of the Austenite had Transformed; (1000×). *(Vilella)*.

In another instance a 1.0 per cent carbon steel showed the following behavior:

Approximate Cooling Rate: 1400° F. to Transformation	Temperature of Transformation	Brinell Hardness Number
3° per minute	1300° F.	217
100° per minute	1260° F.	262
400° per minute	1245° F.	302

With still more rapid cooling the fine lamellar structure ceases to form, and with even 1.1 per cent carbon the hard-
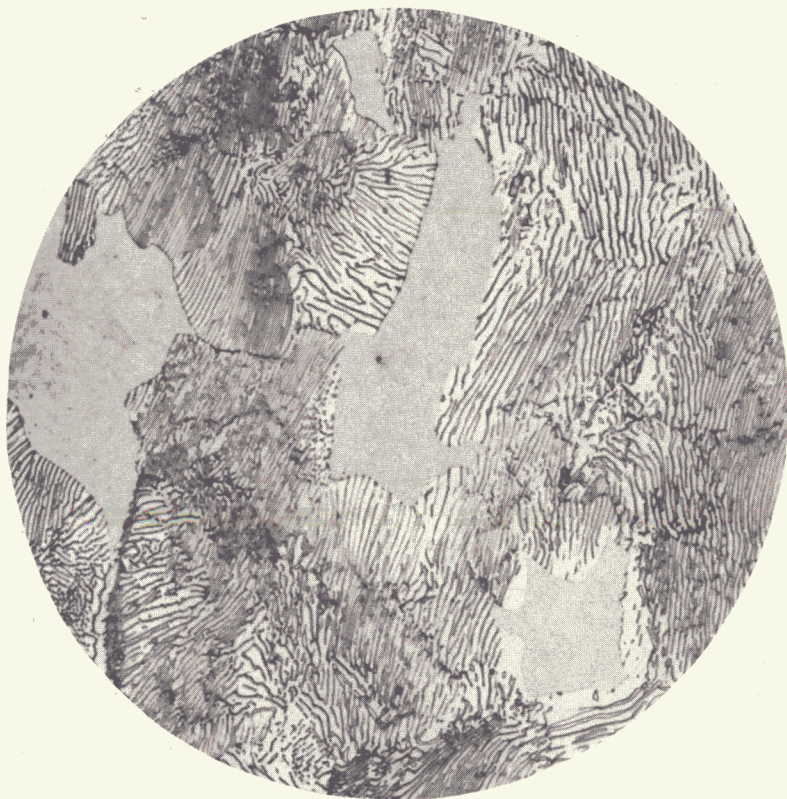
Fig. 18—The Transformation to Pearlite Arrested when 75 Per Cent of the Austenite had Transformed; (1000×). *(Vilella)*.

ness of the lamellar structure is only about 444 to 461 at this limiting condition.   As Mehl[3] so aptly points out, at the rate at which the lamellae are becoming thinner with lower temperature of formation, a temperature must soon be reached at which each lamella would be of smaller dimension than the supposed crystal unit of cementite.   Figs. 20 to 24 show photomicrographs of these lamellar structures for different temperatures of formation, each having the hardness indi-

[3]"The Physics of Hardenability.   The Mechanism and the Rate of Decomposition of Austenite", by Robert F. Mehl. Symposium on Hardenability of Alloy Steels, 1938, American Society for Metals, p. 40-41.

Fig. 19—The Transformation, Shown in Figs. 15 to 18 Inclusive, when all Austenite had just Transformed to Pearlite; (1000×). *(Vilella).*

cated. Considering the cooling rate on the hot-bed of sections about ½ to ¾ inch thick, the hardness varies with carbon content about as shown in the middle curve of Fig. 26.

When the rate of cooling is too rapid for the lamellar type of transformation at about 1000 degrees Fahr., or 550 degrees Cent., the austenite does not transform until a very much lower temperature is reached, assuming the cooling medium itself to be at a temperature below about 200 degrees Fahr. (100 degrees Cent.); the transformation is then of a different sort and a much harder constituent, martensite, is
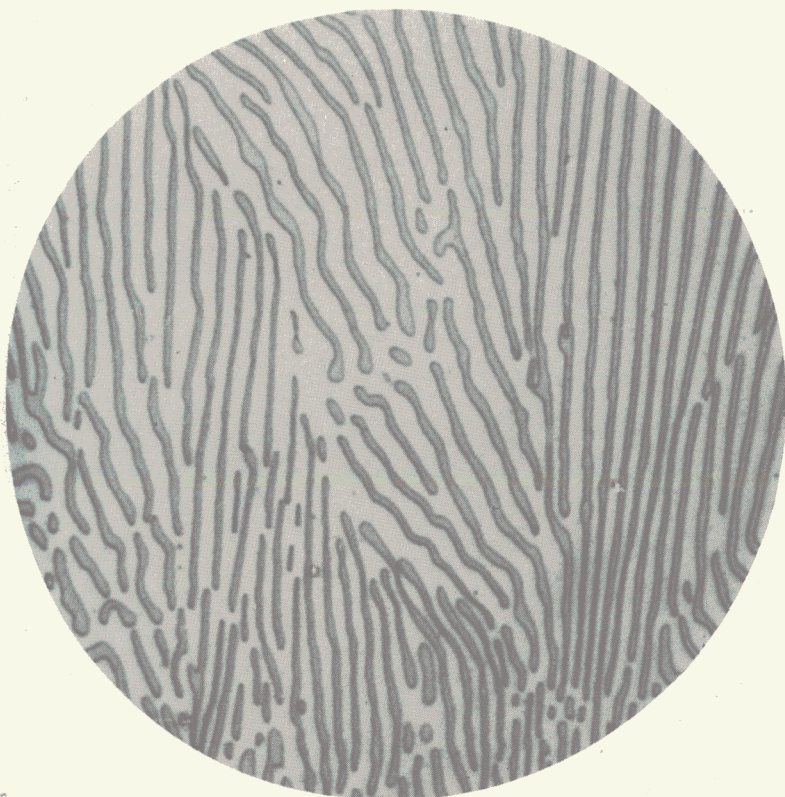
Fig. 20—Coarse Pearlite Formed at about 1325 Degrees Fahr. (720 Degrees Cent.). Hardness R*c* 5, BHN 170; (2500×). *(Vilella)*.

formed. The rate of cooling which just prevents the reaction in the vicinity of 1000 degrees Fahr. and results in the lower temperature reaction is called the "critical cooling rate" or critical quenching rate. The former transformation and its temperature is spoken of as Ar' and the second as Ar''; the critical rate is the cooling rate separating the two. Many steels, however, cooled at about the critical rate, transform in part by one mode and in part by the other, even in the same grain.

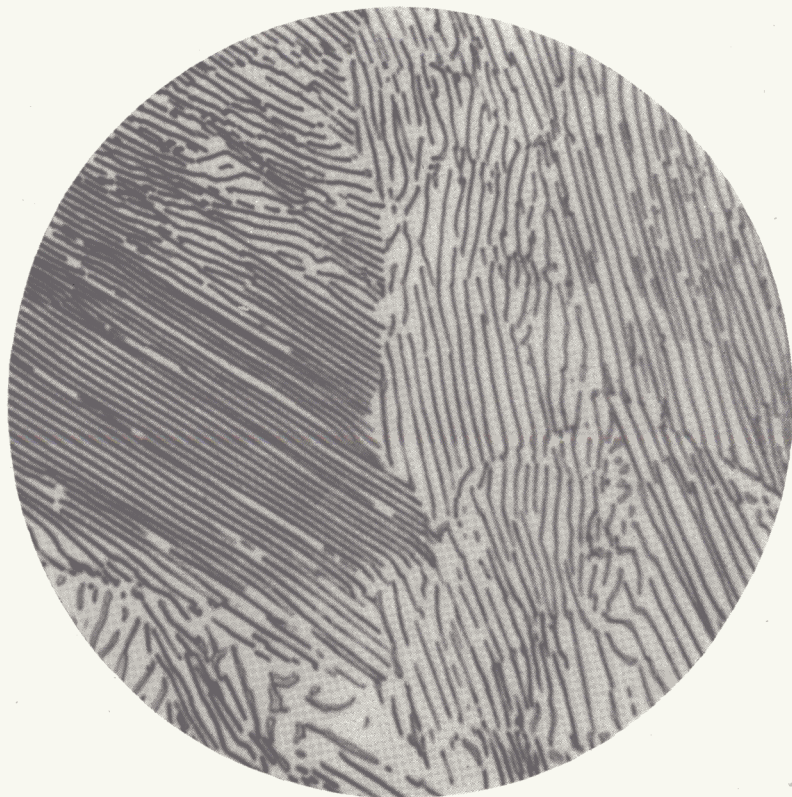Extremely slow cooling of many of the steels under dis-

Fig. 21—Coarse Pearlite Formed at about 1300 Degrees Fahr. (705 Degrees Cent.). Hardness Rc 15, BHN 208; (2500×). *(Vilella).*

cussion may not yield the well-formed lamellar structure at all, particularly if they be low in all alloying elements or if the austenite is fine-grained or if some carbide was left undissolved in heating. Instead, the carbide is partly spheroidized, or somewhat cellular, i.e., at grain boundaries. For this reason it is incorrect or misleading to show in a diagram for the *general* case a transformation, *at equilibrium temperature,* of austenite to lamellar pearlite, for "pearlite" was the name suggested by the pearly appearance of the lamellar association.
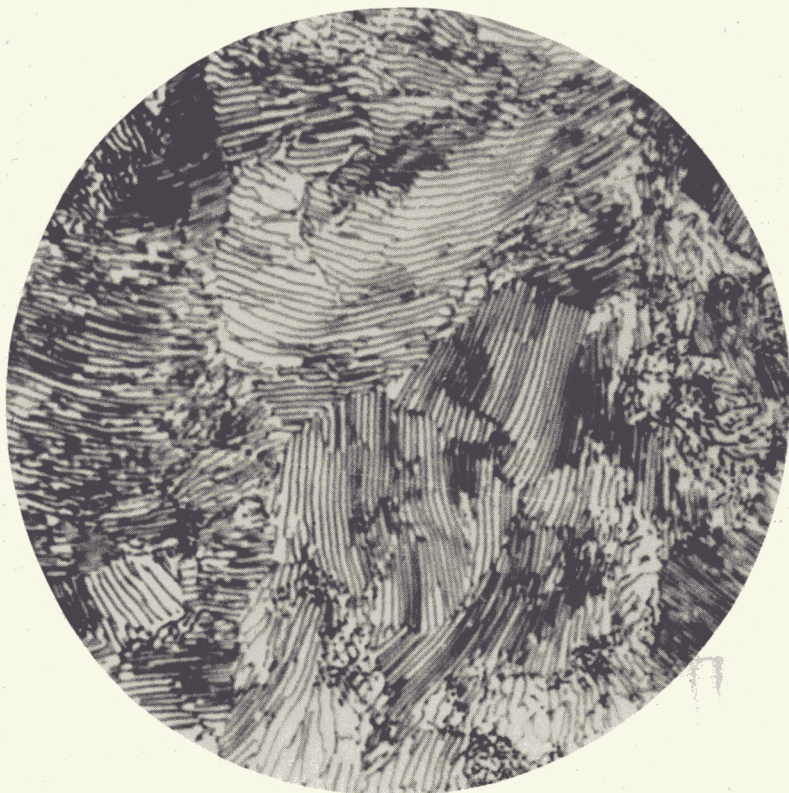
Fig. 22—Medium Pearlite Formed at about 1225 Degrees Fahr. (665 Degrees Cent.). Hardness Rc 30, BHN 296; (2500×). *(Vilella).*

The Pro-Eutectoid Reaction — For compositions somewhat removed from the eutectoid composition, (that is, the composition which transforms wholly to austenite at the lowest possible heating temperature) a rejection of ferrite or carbide may occur before the austenite, thus restricted in carbon, transforms to the lamellar structure. The maximum possible amount of such pro-eutectoid constituent is limited by the composition, but the actual proportion is controlled by the rate of cooling. The higher the rate of cooling the smaller the amount of the pro-eutectoid ferrite or carbide to be found.
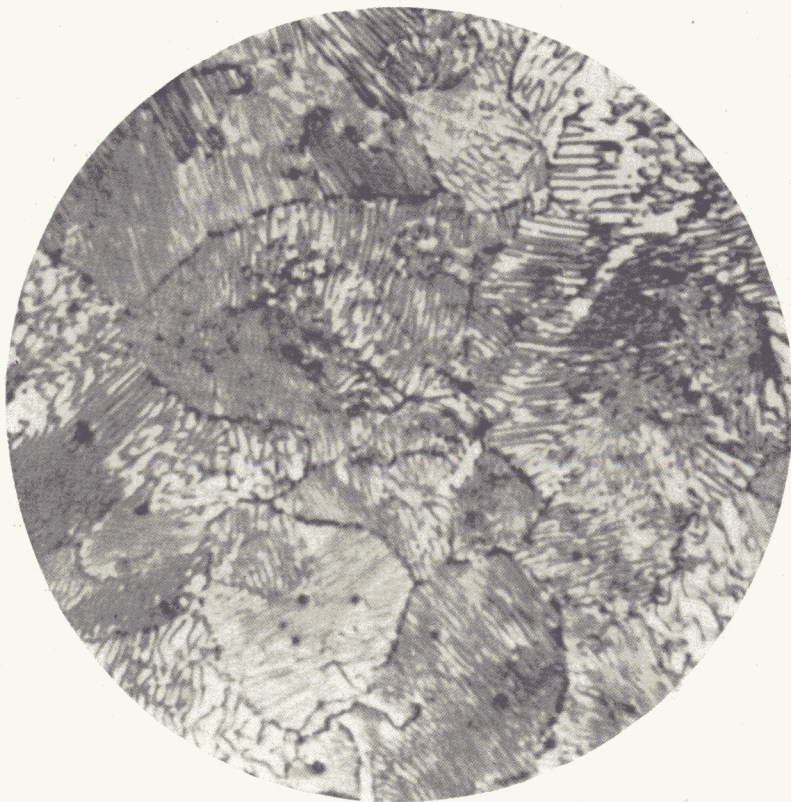
Fig. 23—Fine Pearlite Formed at about 1200 Degrees Fahr. (650 Degrees Cent.). Hardness Rc 33, BHN 319; (3000×). *(Vilella)*.

Thus a 0.50 per cent carbon steel is readily cooled so that only a few per cent of ferrite separates before the remaining austenite transforms to a fine lamellar structure. Somewhat the same state of affairs exists for the hyper-eutectoid steels with respect to carbide, which is, however, more likely to separate than ferrite for an equal remoteness in composition from the eutectoid. Both constituents in non-eutectoid steels form in the grain boundaries of the austenite but ferrite becomes somewhat more irregular than the carbide envelopes.

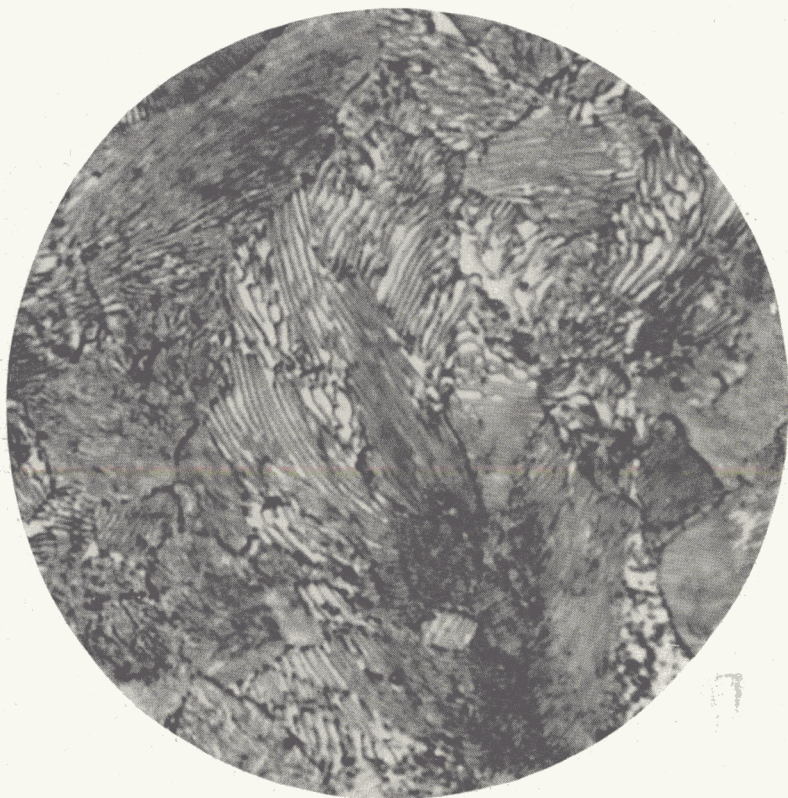The pro-eutectoid reaction is essentially one of diffusion.

Fig. 24—Fine Pearlite Formed at about 1100 Degrees Fahr. (595 Degrees Cent.). Hardness Rc 40, BHN 377; (3000×). *(Vilella)*.

Sizable regions of ferrite alone cannot form except carbon can migrate therefrom and such diffusivity is a function of temperature. Thus the greater the degree of undercooling the less the size of ferrite or carbide masses formed; considering that the entire time within the range at which any ferrite forms is often only a matter of minutes, it is not surprising that equilibrium with respect to the pro-eutectoid reaction is, in most instances, not reached. An interesting circumstance is that some pro-eutectoid ferrite forms in hypoeutectoid steel even when the austenite is brought unchanged
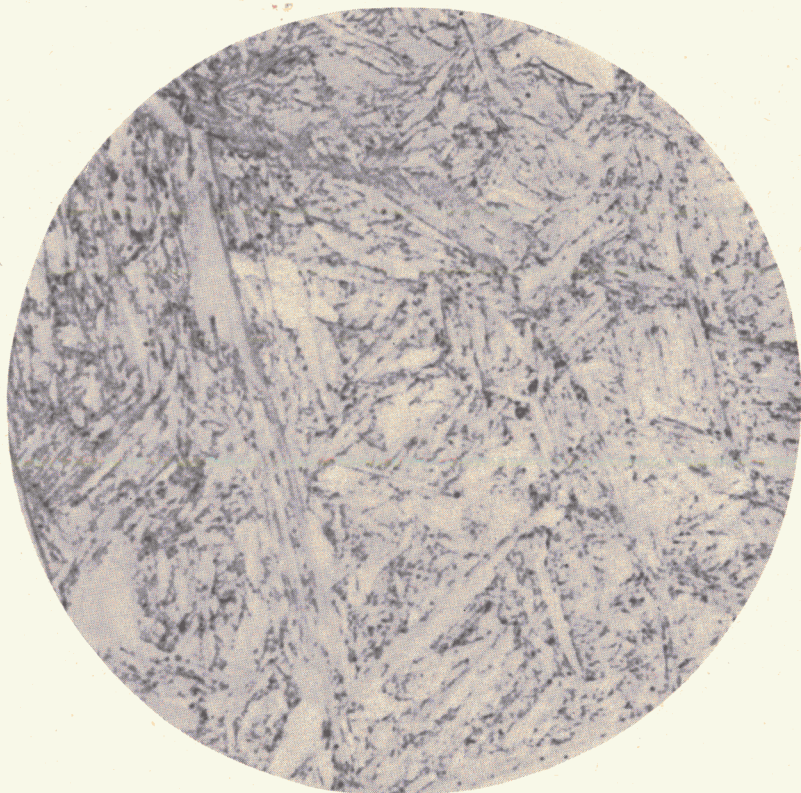
Fig. 25—The Microscopic Appearance of Martensite in Eutec-
toid Steel; (2000×). (Vilella).

to temperatures where austenite is *wholly* unstable. It is as
though the supersaturation aspect were still a controlling one
even when the solvent itself is unstable.

The Acicular Mode of Transformation — When the
carbon steels are quenched at a rate exceeding the critical
cooling rate in a medium near room temperature, the trans-
formation is rapid but it occurs at a temperature in the vicin-
ity of 200 degrees Fahr. (100 degrees Cent.) ; the product is
relatively exceedingly hard and more voluminous than the
other structures. This structure is called martensite, and is

illustrated in the photomicrograph of Fig. 25. Microscopically it has a characteristic acicular structure, — a reflection of its mode of formation, that of successive, nearly instantaneous, transformations on crystallographic planes of the austenite crystal. The transformation, not involving diffusion, almost certainly occurs by the so-called "shear" mechanism; the orientations in adjacent plates from the same austenite grain are not identical since the transformation may proceed upon different austenite planes and since the remaining, as yet untransformed, austenite must suffer some deformation. The martensite is therefore equivalent in part to a fairly fine, flat-grained, strained metal. Its hardness can scarcely depend upon this circumstance, however.

The Nature of Martensite — As to atomic arrangement in martensite, it is basically that of ferrite with a degree of asymmetry corresponding to the tetragonal type which increases with carbon content up to a ratio of about 1.06. It is supposed that the carbon atom, unable at these low temperatures to form iron carbide is responsible for this tetragonality. At any rate, warming to about 100 degrees Cent. restores the isometry of the cubic ferrite lattice without any loss in hardness. We infer that it is possible, up to this temperature, for the carbon to inaugurate a state of affairs resembling the single simple molecule of iron carbide; that is, in its first effort toward initiating carbide crystallites. At any rate the tetragonal state is intermediate and unnecessary to great hardness; the plate-like structure may account for some of the hardness of very low carbon martensite.

It seems that one of the simplest views to take of martensite is that it is a unique example of precipitation hardness in which (1) a potentially phenomenal effective supersaturation is possible by virtue of the allotropy of iron, and (2) the condition of approximately optimum dispersion, that of the smallest conceivable unit of precipitating compound, is
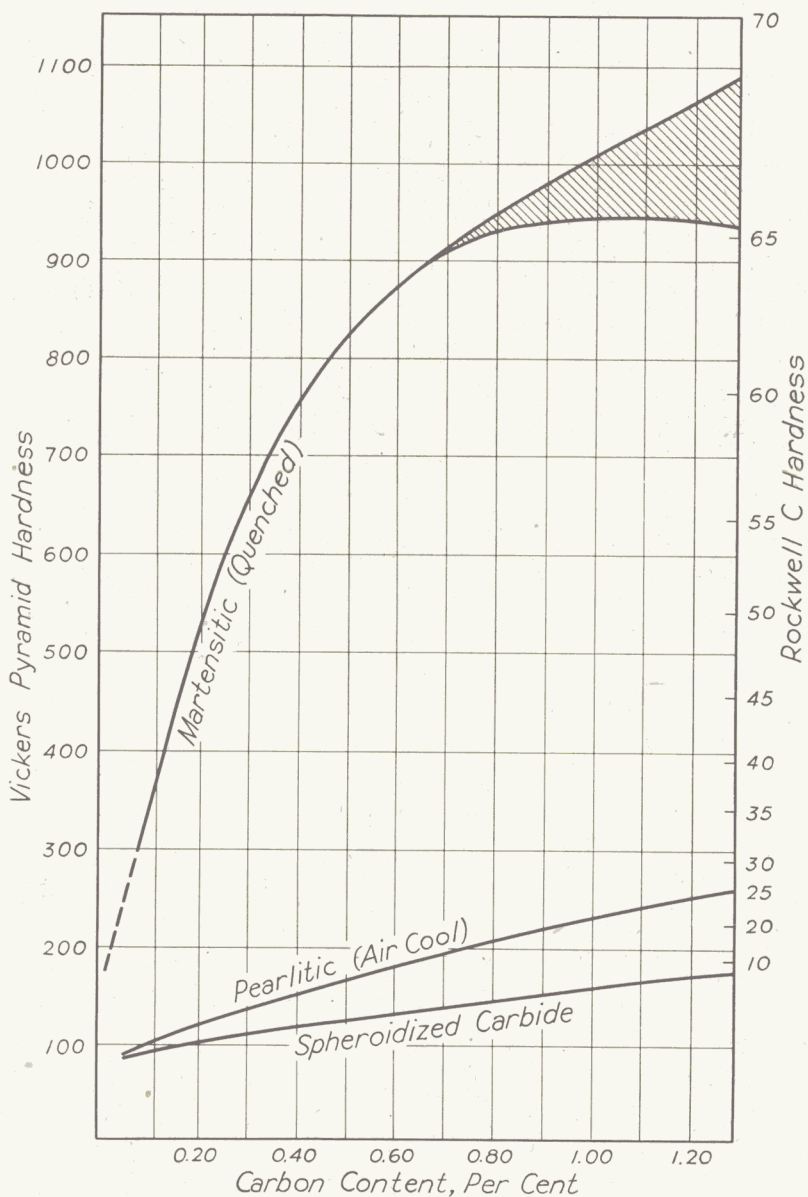
Fig. 26—The Hardness of Carbon Steels in Three Structural Conditions as a Function of Carbon Content: (Approximate).
Top Curve—Fully-Hardened Martensite.
Middle Curve—Lamellar Structure (with Proeutectoid Constituents) as Formed in Normal Cooling after Rolling.
Bottom Curve—Carbide Coarsely Spheroidized. Minimum Hardness Commercially Obtainable.

brought about alone by the transformation induced in quench-ing, without any time-consuming aging treatment. In mar-tensite, the initial precipitation hardness is probably nearly a *true* maximum rather than the purely *statistical* one usually found in precipitation systems, which has incidentally been attributed, probably erroneously, to a so-called "critical size" of dispersed particle. To associate the maximum age-hard-ness with a supposed optimum particle size would be to assume a fair uniformity of particle size. This is probably not realized, since the broad flat maximum of precipitation hardening curves generally suggests that some regions are already softening before others are aged at all. At any rate martensite in carbon steel parallels ideally one aspect of pre-cipitation phenomena in that it softens uniformly with increased heating temperature or prolonged time of heating, during which, particle growth may occur.

The relation of the carbon content to the full hardness of martensite is shown in Fig. 26. The high carbon region of this curve is shown broadened. Due to an increasing tend-ency for some of the softer austenite to be retained perma-nently through the quench, high carbon steels often show a reduced hardness. When quenched, however, with some car-bide undissolved, they reflect the hardness effect of residual hard particles as well as that of the more nearly pure mar-tensite. The highest figures in the range represent a probable maximum hardness of the quenched steels, in the approach to which the metal may be refrigerated to induce further trans-formation of any retained austenite.

Tempering — The softening of the martensite of carbon steels brought about by reheating for 1 hour at increasing temperatures is shown in Fig. 27. This pattern of softening is so definite and universal for the carbon steels as to be almost a kind of characteristic of them. By reheating the martensite, the originally "molecularly-sized" carbide units
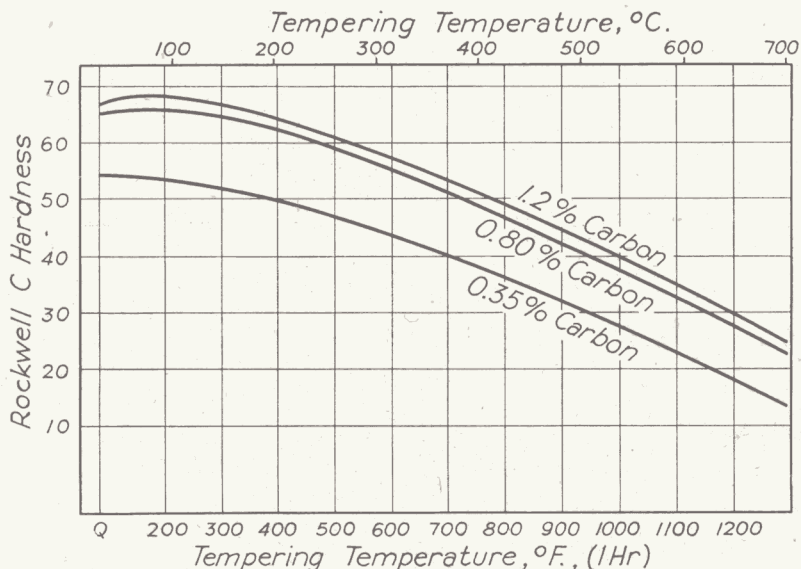
Fig. 27—The Softening of Quenched Carbon Steels by Tempering for 1 Hour at Indicated Temperatures.

grow in size, decrease in number and become, thereby, less and less effective as a hardening dispersion. While there is no irrefragable proof that these particles at submicroscopic size are spheroidal, the overwhelming persuasion is that they are, throughout the entire series. At any rate when the tempering results in a substantial softening, the individual particles of carbide are observed to be roughly spheroidal.

Spheroidal versus Lamellar Structures — It will be apparent that two means exist for securing final hardness between some 190 and 400 B.H.N. in the higher carbon steel. The hardness may be developed by cooling the appropriately heated, and substantially austenitic, steel at a rate adjusted to produce a lamellar structure of the desired fineness and hardness; or the heated steel may be quenched to form hard martensite and then tempered to develop an exactly similar penetration-hardness number. It might be supposed that
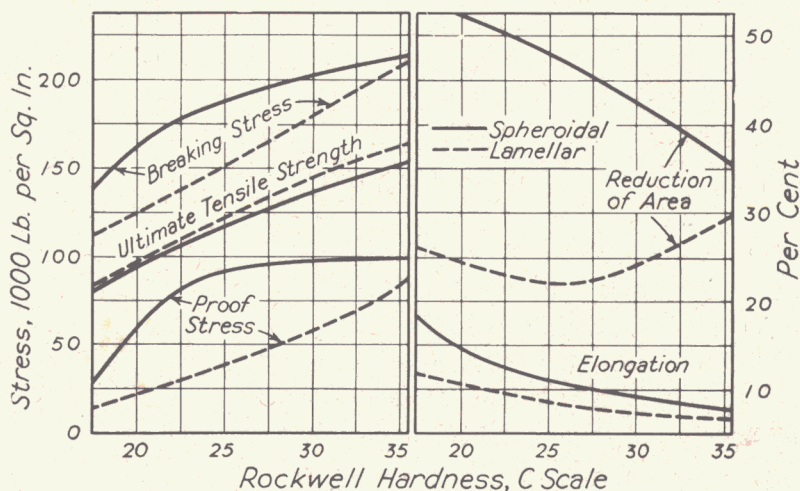
Fig. 28—The Influence of Structure upon Tensile Properties of a Eutectoid Steel at Various Identical Hardnesses as Between (a) the Lamellar Structures (Pearlite) and (b) the Spheroidal Structures (Heat-Treated).

there would be no great difference in the pieces at the same hardness regardless of the heat-treatment selected, but actually the mechanical properties are quite different depending upon whether the carbide is in the form of lamellae or in spheroids. This is shown in Fig. 28, which compiles the results of a series of specimens heat-treated just as outlined above. This carbon steel contains 0.84 per cent carbon, 0.20 per cent silicon and 0.23 per cent manganese.

It will be noted that, for any hardness developed within the range, the plasticity values of the tensile test are superior for the spheroidized structures, that is, for the quenched and tempered steel. Only the tensile strength is lower for the quenched and tempered steel, or as is commonly said, "the heat-treated" steel. This superior strength in the pearlitic series is a reflection of the greater cross-section at the beginning of local elongation and the termination of general elongation, for the reduction of area is far greater for the tempered

steel. The significance of greater reduction of area is this: at the same hardness the steel with the spheroidal carbide dispersion is still capable of greater plastic flow even under the adversity of its considerably greater notch effect at the zone of local constriction. This property is highly valued in many applications. It should be noted that the numerical value of elongation is low; this is due to the circumstance that the length : diameter ratio is not standard but instead is about 10:1.

This chart of Fig. 28 is of some importance in that it shows in a quantitative way the value, in terms of certain engineering criteria, of heat-treatment to develop the carbide dispersion of tempered martensite. If a steel should be incapable of acquiring these superior structures in any large section it would manifest a serious shortcoming and, as will be developed, this is a major deficiency of carbon steel.

Transformation of Austenite at Constant Temperatures — Thus far the structures in carbon steel have been classified as:

    1.   Lamellar (pearlitic).

    2.   Martensitic and its spheroidal derivatives.

The first family of structures in carbon steels forms above about 1000 degrees Fahr. (550 degrees Cent.) upon a not too rapid cooling of the properly heated and austenitic steel.

The second family results from rapid cooling and a transformation not far above room temperature, followed by such reheating as is desired for a given hardness or for a given ductility.

The transformation of austenite can also be studied in another way. Very small specimens may be almost instantly cooled to any temperature before the austenite transforms and then the transformation at this *constant* temperature can be closely followed and its progress and rate recorded. It is

found that in carbon steels the lamellar family of structures, as expected, is produced at constant holding temperatures above about 1000 degrees Fahr. (550 degrees Cent.) and that martensite quickly forms in the range of temperature below about 300 degrees Fahr. (150 degrees Cent.). What of the temperatures lying between? The austenite transforms here also, although in many carbon steels one could never become aware of it if the cooling mediums were always below, at, or even about room temperature regardless of the heat-abstracting power. The product is acicular with the spheroidal type of carbide particle, and during transformation the formation of this aggregate occurs apparently by successive crystal-plane or block transformations as in martensite, rather than by nodular growth (at least in the lower temperature range).

The rate of transformation at different temperatures is interesting and significant. Fig. 29 shows the transformation rates in a schematic way for a eutectoid steel: austenite (black) transforms to ferrite and carbide or martensite (white) in accord with the time interval plotted on a logarithmic scale. The pattern of rates as they change with transformation temperature readily explains the reason for the Ar' and Ar'' temperatures as they vary with cooling rate. The photomicrographs of Figs. 30, 31 and 32 illustrate the range of structures formed in the intermediate division of possible transformation temperatures. These structures known as "bainite" formed at temperatures roughly between 1000 and 300 degrees Fahr. (550 and 150 degrees Cent.) in carbon steel have interesting mechanical properties, often tougher for the same hardness than tempered martensite.

Fig. 29 indicates the beginning and end of the transformation of austenite to the ferrite-carbide families of structures insofar as the reaction proceeds at a measurable velocity. In the temperature range around 350 degrees Fahr. (175 degrees Cent.) all indications of reaction cease in a few
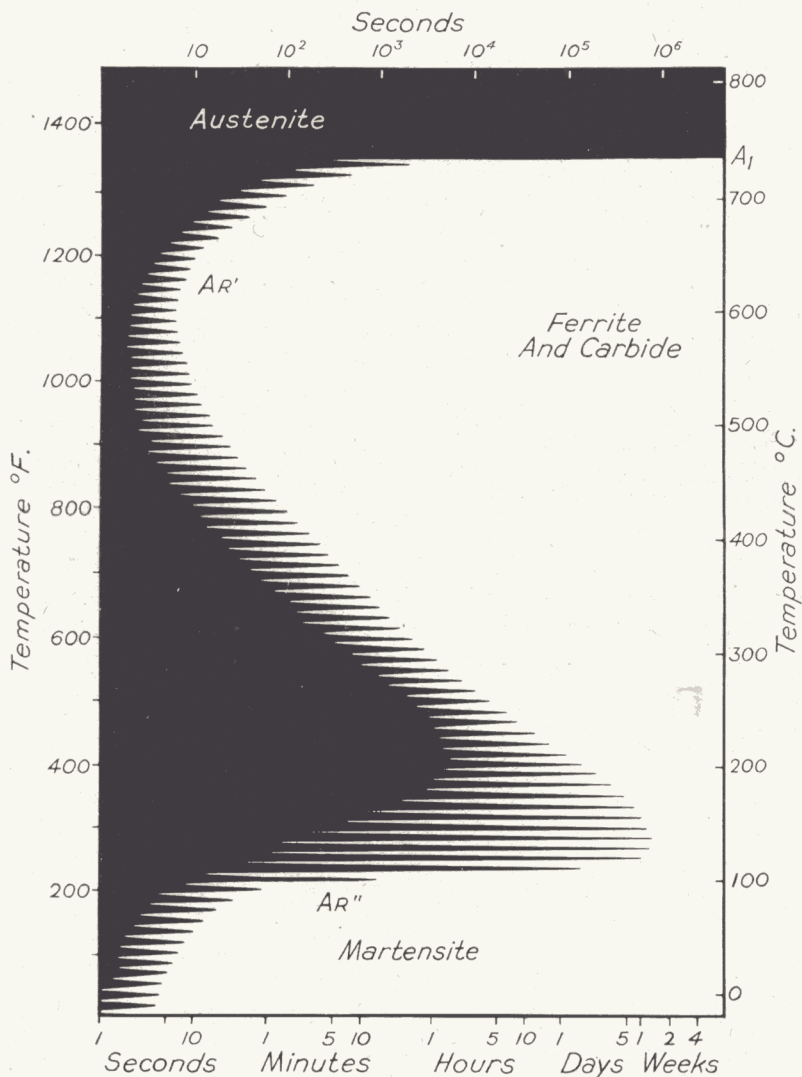
Fig. 29—The Time Interval Required for Transformation of Austenite (Black) to Ferrite and Carbide (White) at Constant Temperatures. Based upon a Eutectoid Steel, Schematic.

weeks, and below this temperature the reaction halts much sooner. But in the steels carrying more than some 0.70 per cent carbon a little austenite may still be present and remain

Fig. 30—Typical Microscopic Appearance of Acicular Structure as Developed by Transformation at 925 Degrees Fahr. (495 Degrees Cent.) just Below the Temperature for Maximum Rate. Note Trace of Nodular Fine Pearlite with Bainite; (2500×). *(Vilella)*.

unchanged indefinitely. The diagram of Fig. 29 does not indicate this, since it concerns itself with the reaction alone, as it actually begins and ceases. Accordingly from the physical-chemical viewpoint, the phase austenite may not be regarded as disappearing entirely and the diagram of Fig. 34 is therefore introduced to suit the second viewpoint, as applied to a eutectoid steel which may retain up to about 20 to 25 per cent residual austenite. The hardness of the structures resulting from direct, constant-temperature transfor-

Fig. 31—Typical Microscopic Appearance of Acicular Structures Developed by Transformation at about 550 Degrees Fahr. (290 Degrees Cent.). Bainite; (2500×). (*Vilella*).

mation is shown for a eutectoid carbon steel in Fig. 33.

Hardenability — Referring to the S-shaped curve (Fig. 29 or 34) which summarizes the rates of austenite decomposition, it will be seen that to achieve full martensitic hardness a piece of steel must cool too quickly to be caught in the rapidly-acting tendency toward transformation at about 1000 degrees Fahr. (550 degrees Cent.). It must escape the reaction at Ar′ at its greatest velocity. Obviously if that reaction can be retarded then the cooling rate may be slower. In other words, the critical cooling rate is largely a function of the maximum

Fig. 32—Typical Microscopic Appearance of Acicular Struc-
ture as Developed by Transformation at about 350 Degrees Fahr.
(180 Degrees Cent.). (2500×). *(Vilella).*

rate of onset for the austenite → ferrite-carbide (lamellar)
reaction. This is in turn a matter of the maximum nucleation
rate for the formation of ferrite or carbide nuclei. As it hap-
pens, in most steels these nuclei are located largely in the aus-
tenitic grain boundaries and, therefore, the more grain
boundary surface the more nuclei. Any change in the steel
which delays and retards the transformation of the austenite
in the range of its maximum Ar′ rate contributes to slower
critical quenching speed and in turn to its *hardenability.* Such
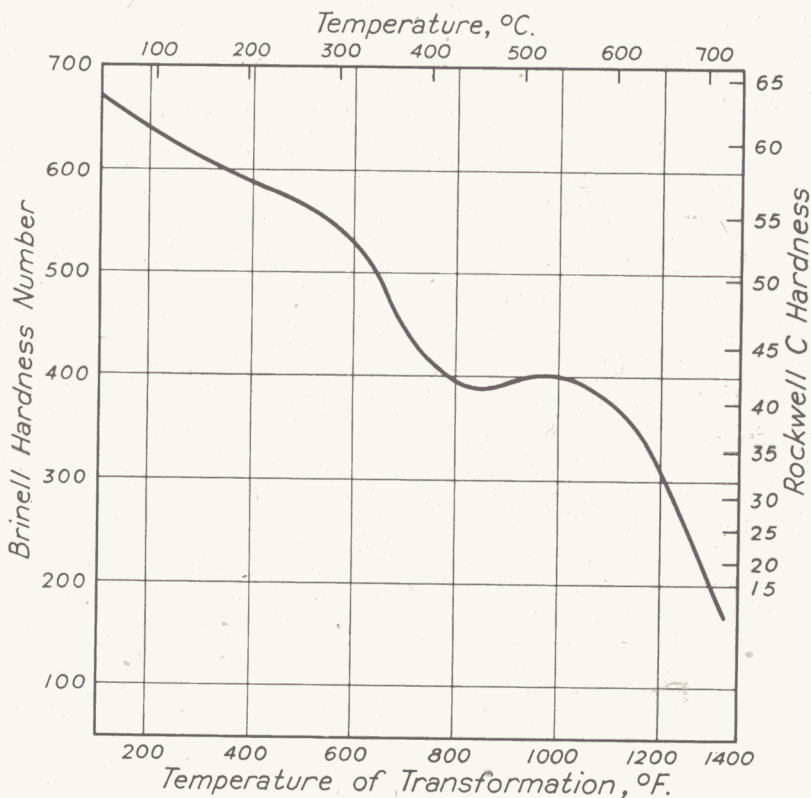changes as increase hardenability do not necessarily have any

Fig. 33—The Hardness of the Products of Austenite Transformation at Constant Temperature as a Function of the Transformation Temperature, Eutectoid Steel. (*United States Steel Corporation Research Laboratory*).

effect whatever upon the maximum martensitic hardness to which a steel may be hardened; maximum intrinsic martensitic hardness is quite another matter and is more influenced by carbon content than by any other factor. Arbitrarily perhaps, it is here elected to define hardenability as the capacity of a steel to avoid transformation to soft products with a lower and lower cooling rate, regardless of just what maximum hardness it may achieve in the martensitic state. Thus hardenability, which relates to the minimum quenching rate which
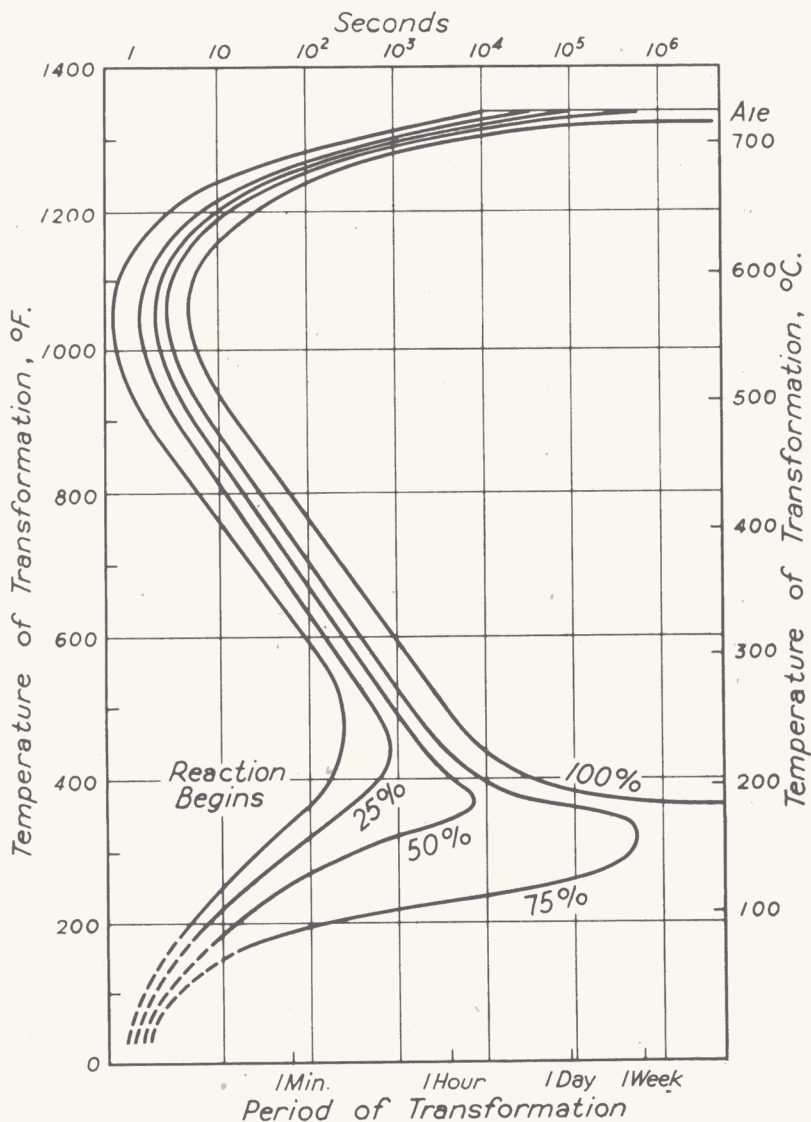
Fig. 34—The Time Interval Required for 25 Per Cent, 50 Per Cent, 75 Per Cent and 100 Per Cent Transformation of Austenite at Constant Temperature in a Eutectoid Steel. Note that Although Reaction apparently reaches a Standstill it is shown as Actually Incomplete. This is Scientifically more Acceptable than Fig. 29.
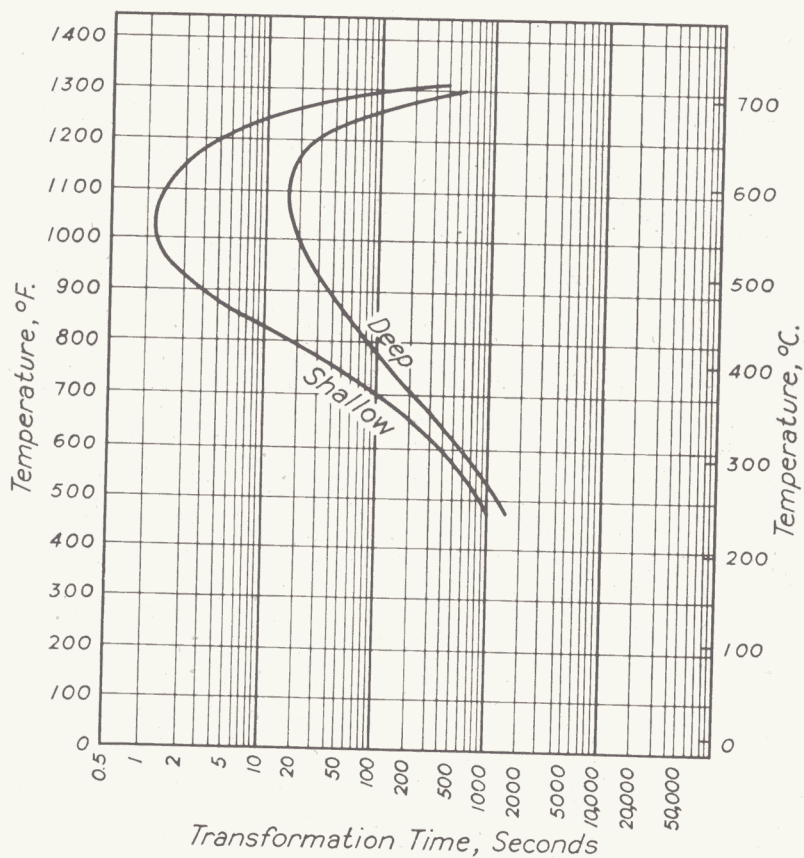
Fig. 35—Time Interval for 50 Per Cent Transformation at Temperatures Significant in Hardenability for a Shallow and a Deep Hardening Low-Alloy Steel.

will produce martensite and permit the steel to escape transformation to softer constituents, controls the size of section which may be uniformly heat-treated to initial high hardness and controls also the depth of hardening in larger sections of homogeneous chemical composition. Fig. 35 shows the transformation rates of (1) a shallow hardening steel in comparison with the rates for (2) a steel of deep hardening character.

Grain Size and Hardenability — It has been mentioned

that the austenite in the furnace ready for quenching or other controlled cooling has two characteristics of prime importance:

1. Composition and homogeneity of composition.
2. Nucleation capacity or effective grain size.

Assuming for the moment substantially homogeneous austenite of carbon steels (not so easy to achieve), let us discover the effect of the austenitic grain size. Austenite, following the familiar pattern of grain growth in most metals, coarsens with higher heating temperature. Depending upon composition, inclusions, etc., this growth of the grains may be gradual or more or less abrupt but heating temperature provides a means of controlling to some extent the grain size of the austenite. Identical test pieces may be individually heated to develop a range of grain size by control of heating temperature. All may then be quenched into the same large bath by identical manipulation and the influence on hardenability may be explored. An exceedingly small difference in actual cooling rate results alone from the variation of the initial temperature of the piece at the beginning of the quench, but even this discrepancy can be avoided by cooling all the pieces to the same temperature prior to quenching, the maximum temperature previously reached having been different for each. Such specimens of suitable, identical dimension, ready to be identically quenched, are well adapted for estimation of hardenability as influenced by grain size.

Since heat is abstracted at the surface and may flow only from hotter to cooler zones, it follows that the interior of a piece must, during the early and significant part of the cooling period, lose temperature more slowly than the exterior. There are represented then, in one sizable quenched piece, such as a cylinder, a variety of cooling rates. These have been systematically studied and are therefore well known with respect to distribution of rates. If the piece is large enough

Temp. °F.:	A 1800	B 1700	C 1575	D 1450	E 1375
°C.:	980	925	855	785	745
G.S. (mean)	2	3	3.5	4	5

C 0.74
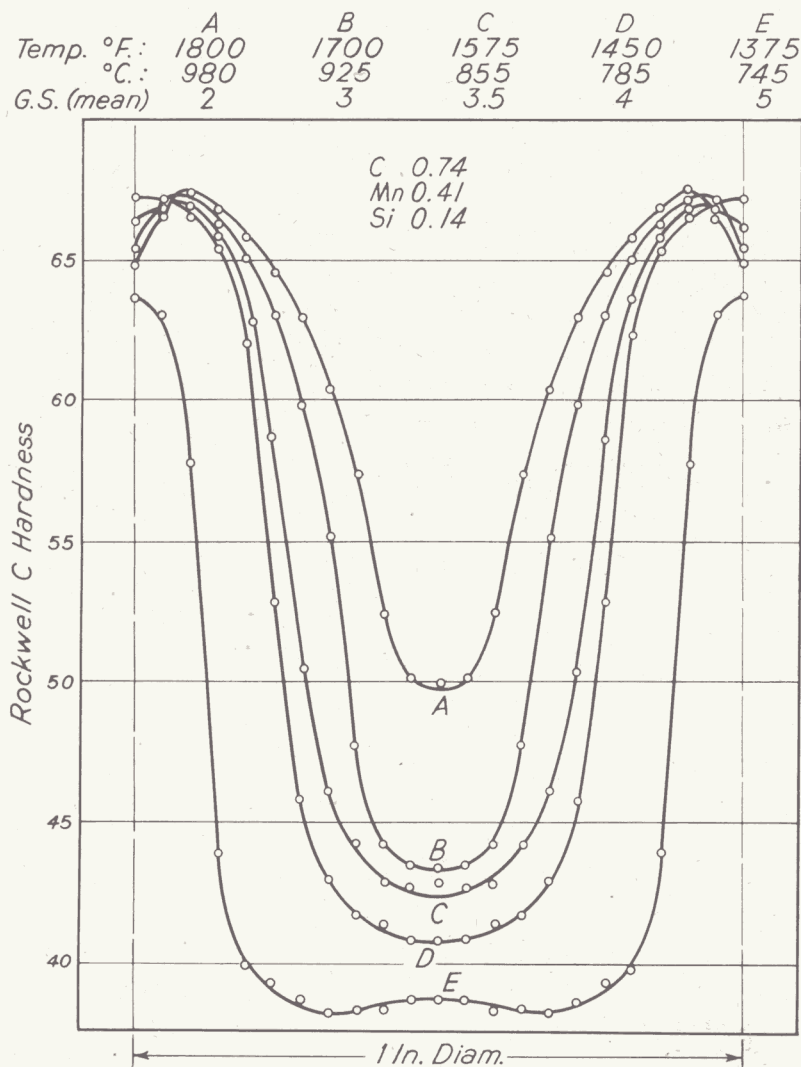Mn 0.41
Si 0.14

Rockwell C Hardness

I In. Diam.

Fig. 36—Influence of Austenitic Grain Size upon the Hardenability of a 1-Inch Round of 0.75 Per Cent Carbon Steel. Quenching Temperature Identical for all Grain Sizes which were Induced by Variety of Maximum Heating Temperatures as Shown.

some interior zone must, in the quench, cool more slowly than the critical quenching speed while the exterior transforms to

hard martensite. In the case of carbon steels, a 1-inch round is frequently satisfactory for securing an unhardened interior with water-quenching.

Fig. 36 shows the distribution of hardness across a representative diameter of 1-inch rounds of a 0.75 per cent carbon steel with a variety of austenite grain sizes established prior to water quench. The influence of coarser grain to increase hardenability is clearly shown, corroborating the circumstance that the critical cooling rate for hardening is decreased by an increase in austenite grain diameter. In this instance all specimens were homogenized by a prior heating and a rapid subcritical cooling rate and all were quenched from the same temperature; all being taken from the same bar of steel, there is no question of any significant influence other than nucleation capacity as reflected in grain size. It should be stressed that for a given composition the nucleation capacity is the actual controlling factor in hardenability but that in a great many steels this capacity appears to be surprisingly well correlated with the true austenitic grain diameter. In connection with the subsequent discussion of alloying elements this matter will be considered in greater detail.

It will be understood that a certain degree of hardenability is a characteristic of any austenite depending upon its composition and its grain size; or more scientifically, a characteristic of austenite depending upon its homogeneity, its composition and its nucleation capacity. High hardenability, however, is not a universally desired characteristic; there are many applications for shallow hardening steel in which the softer unhardened interior is an advantage. Whenever it is necessary to develop optimum combinations of high strength and ductility throughout a large section, then high hardenability is a requisite. In principle, higher hardenability may be secured through the achievement of a very coarse grain size in the austenite, but this method may not be safely
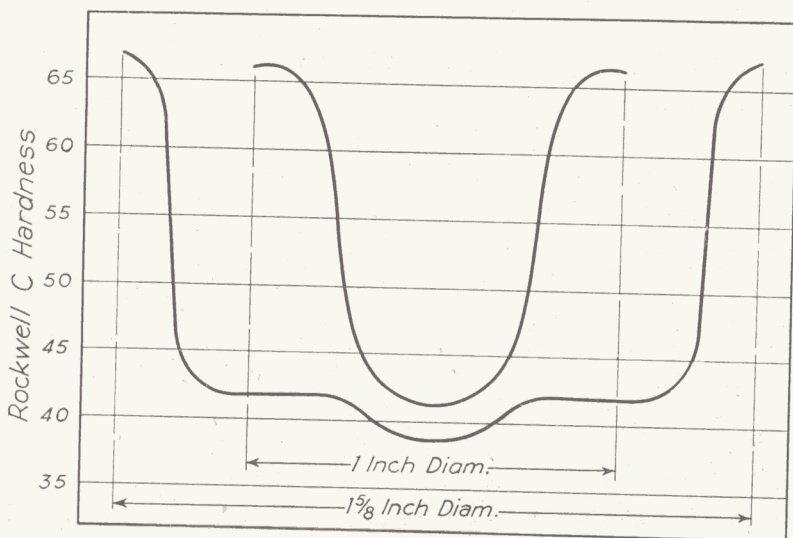
Fig. 37—The Distribution of Hardness across Representative Diameter in a Typical Eutectoid Steel of Medium Grain Size in 1-Inch and 1⅝-Inch Round Sections. This is an average Degree of Hardenability as reflected in Warm Brine Quench.

utilized to any great extent. Rather, hardenability must be developed by a change in the composition of the austenite and for this purpose the alloying elements are effectively employed.

Limitations of Carbon Steel in Hardenability — Fig. 37 shows a general degree of hardenability in high carbon steel which may be regarded as good practice in a 1-inch and 1⅝-inch section. This degree of hardness penetration is securable without undue coarsening of the grain and without an extra drastic water quench. The hardened zone is already very thin in the 1⅝-inch round section as water quenched. To secure much deeper hardening by coarsening the grain would be to invite quenching cracks. In the charts of Fig. 38 the approximate maximum hardness penetration, based upon about 45 Rockwell C, is shown as a function of diameter of the round section in 0.45 per cent carbon, 0.75 per cent manganese, forging steel. Some such steels would yield a some-
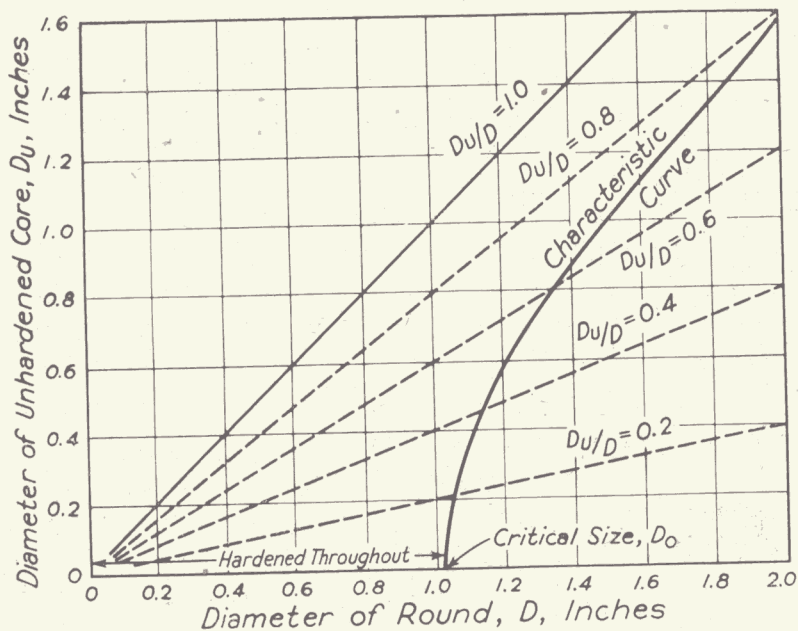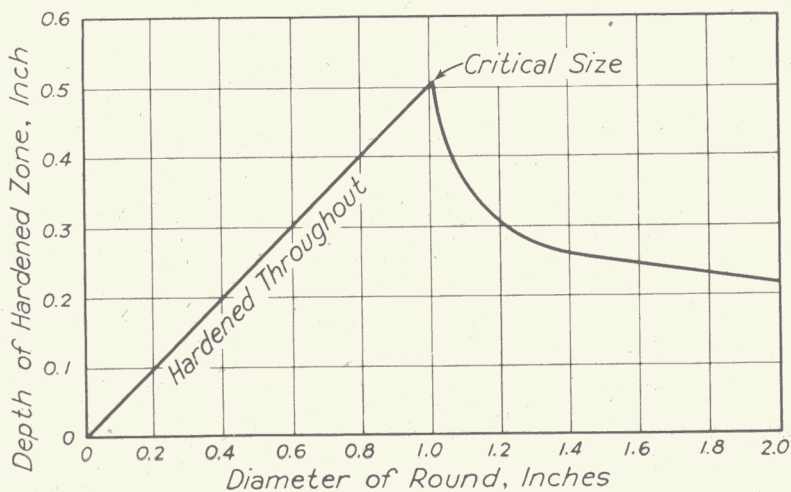
Fig. 38—The Hardenability of a 0.45 Per Cent Carbon Steel, S.A.E. 1045, as presented in Two Charts. The Thickness of Hardened Exterior Rim (Above) and Diameter of Unhardened Interior (Below) are Plotted against Diameter of Round. Vigorous Water Quench. *(Grossmann).*

what lower depth, — high pressure water for quenching might induce a slightly deeper penetration of the 50 Rockwell C zone but the chart illustrates a general average. Much greater hardenability could not be expected in the grade. The degree of hardness penetration sought in materials for many familiar machine designs is far greater than that shown in the foregoing examples. Obviously this cannot be accomplished in carbon steel, and the introduction of alloying elements, which generally enhance hardenability, is resorted to.

For many purposes steel which retains a fine austenite grain but has a high hardenability is desired. The reasons will become apparent.

Austenite Grain Size and Final Properties — It would naturally be expected that a heat-treated shape, fully hardened at the surface, but of sufficient size to have at the center an unhardened zone would be tougher, all in all, than one fully hardened throughout. Fine grains of austenite in one, and coarse grains in the other piece, both of the same composition, could bring about these two conditions; but fine grain, *per se,* would not alone be responsible, for a milder quench with the same grain size would also have provided the tough core. Fine grains exert a much more specific influence than this. Even in small sections, hardened fully throughout, the martensite as formed from the finer grains is somewhat tougher than that from coarse austenite, and this extra toughness is preserved in moderate tempering. Fig. 39 shows how the energy required in breaking small specimens varies with the grain size established in the austenite at heat-treating temperature. In this instance three different heats, having somewhat different coarsening tendencies but of essentially identical composition, heated at three different temperatures, sufficed to develop a variety of grain sizes. Small-sized specimens insured complete hardening, while a tempering treatment produced a hardness of 50 Rockwell C for all, at which
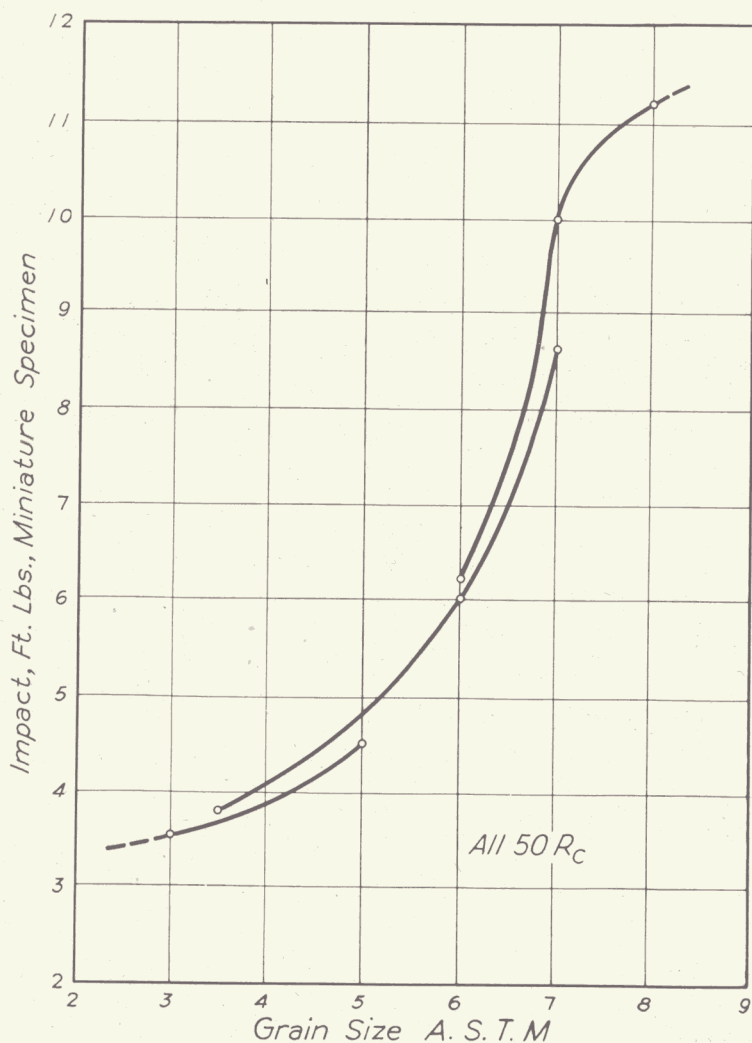
Fig. 39—Relation of Prior Austenite Grain Size to Toughness (Energy absorbed in breaking Small Specimens) of Eutectoid Steel Heat-Treated to 50 Rockwell C.

hardness the energy in breaking is at a highly sensitive range of values. It is somewhat unusual to secure such consistency because other factors often obscure, in part, the pure grain size effect. It is thought that coarse austenite grains build up

and maintain higher internal stress in transformation than fine ones, resulting in a lower net external stress required for the first rupture. In the notched bar test, once a notch is propagated, breaking follows with little additional applied load.

Further effects of austenite grain size are observed in the products of transformation. When the product is lamellar or pearlitic the large grains yield a harder product for the lamellae are more finely spaced, and if the steel is hypoeutectoid the volume of free (pro-eutectoid) ferrite is much less. The pearlitic steels resulting from a coarse prior austenite grain often machine more freely because of the lowered plasticity accompanying the structural effects above.

In hardening, warpage is less when a fine austenite grain is established in the steel as quenched. Probably this is a necessary accompaniment to the decreased tendency for quenching and grinding cracks in the same fine-grained steels, and the accompanying phenomena are probably related to lower stress in the products resulting from the hardening of fine-grained austenite.

SUMMARY OF CARBON STEELS AND NEED FOR ALLOYS

In appropriately small sections carbon steels may, within limits, be cooled according to schedules and rates which suffice to produce almost any of the possible distributions of cementite in the ferrite, including the limiting condition of hard martensite. For any one particular type of carbide distribution, the hardness and strength are a function of carbon content; with a constant carbon content the strength increases with the fineness of the carbide dispersion, while the ductility (plasticity) and associated properties decrease. At equal hardness, however, the spheroidal dispersion possesses greater plasticity (toughness) than the lamellar structures.

The transformation of coarse austenite grains is subject to greater lag than that of fine grains. Coarse-grained austenite therefore hardens more deeply in equivalent quenches; fine-grained austenite warps less on hardening, is freer from internal stress and tougher in the hardened state both after quenching and after moderate tempering.

Relatively small specimens of carbon steel can readily be selected as to composition and heat-treatment to develop practically optimum mechanical properties at room temperature. It may be that with these limitations prescribed, carbon steel is capable of acquiring nearly the full optimum combination of mechanical properties to which a dispersion of hard particles in a matrix of ferrite may attain. The table below illustrates the point with reference to tensile properties:

Property	Commercial Pure Iron	Pearlitic 1% Carbon	Spheroidized Annealed 1% Carbon	Heat-treated 1% Carbon
Yield Point	26,000 psi	85,000 psi	40,000 psi	
Tensile Strength	42,000 psi	150,000 psi	78,000 psi	260,000 psi
Elongation, $\dfrac{L}{\sqrt{a}} = 4.45$	40-44%	10%	31%	13%
Reduction of Area	70-75%	12-15%	57%	45%
Brinell Hardness	80-85	300	156	540

The first two columns show how the presence of carbon alone suffices to impart to the rolled steel a greater strength, taking, however, a considerable toll in ductility. The third column indicates the restoration of some of the ductility by producing the coarsely spheroidized structure with a practical heating schedule. The last column indicates how a structure, intermediate between martensite and the softer, coarser spheroidal one, provides a very high strength with a fair degree of ductility. It is to this range of properties that particular importance attaches, for much alloy steel is so heat-treated as to secure properties in this range, and by comparable control of structure. It has already been emphasized that carbon steel can achieve such properties only in relatively

small section. Nevertheless, from the scientific standpoint the optimum possible combination of room-temperature tensile properties in carbon steel is such as to leave no tremendous difference to explain on the basis of the presence of alloys in comparably small sections. One can at least rest assured that nearly pure iron in which a suitable dispersion of iron carbide can be precipitated offers excellent tensile properties, and that it is not to be expected that the presence of other elements could better these particular properties in a tremendous degree. The alloying elements contrive, however, to enhance certain properties but as will be seen they mainly make possible the practical achievement of some of the optimum combinations of room-temperature properties which, in carbon steel, cannot be realized except in restricted sizes and under restricted conditions.

When the tensile properties at elevated temperature, as well as creep strength, and other special properties such as notch toughness, are under consideration, the contributions of alloying elements become more apparent. Wear resistance is enhanced by special constituents brought about by alloying elements. These advantages, in general, are at a maximum only when the heat-treatment is precisely adjusted to the composition to insure the optimum structure.