

Chapter V

EFFECTS OF THE ELEMENTS IN HARDENING STEEL

It will be recalled that in small sections, carbon steel may be quenched so as to develop the hard structure, martensite, which, upon tempering, yields softer, tougher structures with properties generally superior to those of the equally-hard, but pearlitic, structures of unhardened steel as more slowly cooled. It would be desirable to secure this superior array of properties even though one need be at considerable pains to do so and to secure them not only in small section, or near the surface of moderate section, but throughout large sections also.

Need for Heat-Treatment and Alloys—The presence of alloying elements makes all this (and more) possible and in this respect alone, even though there were no other benefits, the various elements would be well worth their cost. Thus the effects of alloying elements upon hardenability become one of the most important topics of these discussions. In Fig. 28 were shown, in comparison, the properties of quenched and tempered steel and those of pearlitic steel over a range of equal hardnesses. Practically, the prevention of the transformation of austenite to lamellar structures is the primary function of quenching steel, and a most important reason for the use of deep-hardening alloys, even though it is more customary to speak of the hardening of steel with only the realization of the full martensitic hardness in mind. It should be mentioned also that quenching and tempering would probably have replaced the careful selection of cooling rate as a means of adjusting hardness and strength even though the mechanical properties were equivalent at all hardnesses, which they are not. The ease of control through tempering is in all respects superior to that in the cooling rate.

CHARACTERISTICS OF THE AUSTENITE-MARTENSITE
TRANSFORMATION

The intrinsic hardness of martensitic steel is, first of all, a function of its carbon content although there are other factors exerting a smaller influence. The secondary factors warrant some enquiry.

Martensite Formation — The reaction by which austenite transforms to martensite during a quench occurs at an exceedingly high rate at least in the early stages. Moreover it is a reaction which appears to halt at a certain stage depending upon the minimum temperature. Martensite formation in a 0.70 per cent carbon steel makes its appearance at a maximum temperature of perhaps about 400 degrees Fahr. (200 degrees Cent.) and apparently proceeds *only with cooling* down to about 200 degrees Fahr. (100 degrees Cent.) at which point all austenite that is thus to transform has done so. Some regard the transformation as essentially instantaneous at any temperature within the range, i.e., that there is no lag or undercooling. However this may be, the martensite formed by slow cooling between the beginning and end temperatures of martensite formation is a very different material from that rapidly cooled through just the same range; whether this latter be merely a more rapid formation of martensite or a formation at a lower temperature is not clear. The martensite formed by the slower reduction of temperature is not quite so hard, and tougher by far; that formed by the rapid cooling reveals microscopic cracks when the grain is not fine. The superior product is probably generally less stress laden and has long been approached by a brief, closely-timed water quench followed by a final oil-quench.

Retained Austenite — In well heated and hardened steels of carbon content over about 0.55 per cent some austenite is usually persistently retained after the quench particularly in

the presence of nickel, manganese, and chromium in which case the carbon may even be lower. Upon heating to a suitable temperature such retained austenite will transform about in accord with the isothermal transformation pattern for the steel. In some cases refrigeration to low temperature will cause the retained austenite to transform in part, and with the occasional formation of cracks. The presence of some ten or fifteen per cent of austenite with martensite does not apparently greatly reduce the aggregate hardness, and its estimation by X-ray or magnetic techniques is difficult and generally only approximate. Surprisingly little is known about the effects of small quantities of retained austenite; under some conditions, however, the product of its final transformation is not at the optimum of mechanical properties if the drawing temperature is high.

Generally speaking a drastic water quench results in less retained austenite than does an oil quench when either is adequate for the section. Apparently maximum austenite is transformed to martensite when the quench is rapid and continuous down to, or below, room temperature, a condition also likely to set up great stress and to cause cracking.

Alloys and Martensitic Hardness—The next secondary factor in the hardness of martensite is that of composition of the parent austenite in terms of the elements other than carbon. Contrary to general belief the other elements do exert a hardening effect when —

- 1) the carbon content is relatively low,
- 2) the quench is such as to develop maximum hardness.

This will be appreciated when nearly carbon-free steels are considered. Fig. 86²² shows the hardness of well annealed and of quenched iron-chromium alloys carrying about 0.02 per cent carbon. There can be no question of the martensitic

²²The specially rapid quenching of these low carbon materials and their testing was carried out by J. R. Vilella, R. H. Aborn, J. W. Price and E. H. Wyche of the United States Steel Corporation Research Laboratory, and the data on annealed and hardened specimens are gratefully acknowledged.

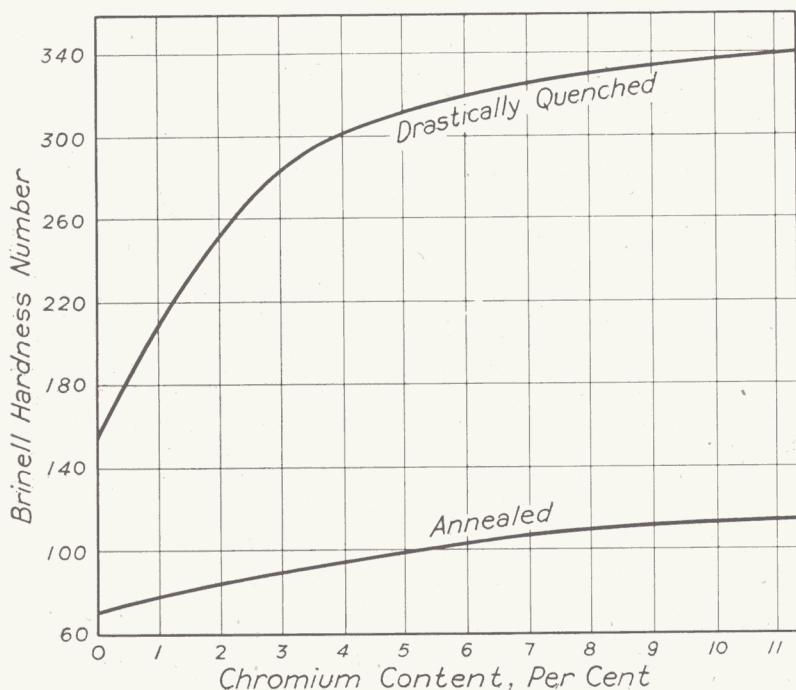


Fig. 86—Martensitic Hardness in a Series of 0.02 Per Cent Carbon, Iron-Chromium Alloys. For reference the Hardness in the Well-Annealed State is shown. (See Footnote 22).

hardness contributed by the chromium in these nearly carbon-free martensites, for the specimens were so thin as to make hardenability itself of no consequence; the quenching rate easily exceeded the critical.

The effect is still observable in 0.35 per cent carbon steels. From a large number of tests²³ upon chromium- and molybdenum-bearing steels of 0.35 per cent carbon content, so quenched for this study as to attain maximum hardness, the increment in hardness is found to be substantially as shown in Fig. 87.

There is some question as to whether or not large pro-

²³The data of Fig. 87 were deduced from the results of an investigation of a number of chromium, molybdenum, and chromium-molybdenum steels carried out by J. R. Vilella, H. A. Brown and R. V. Hilkert of the United States Steel Corporation Research Laboratory. Minor variations in carbon content were compensated for.

portions of alloying elements actually lower the *pure martensitic* hardness below that of a carbon steel of equivalent carbon as dissolved in the austenite. Thus no steels have been reported as hardening more intensely than plain carbon steels. To be sure it is usually not possible to secure quite the initial hardness in quenched cutlery stainless steel that may, at the optimum, be secured in a carbon steel of equivalent carbon.

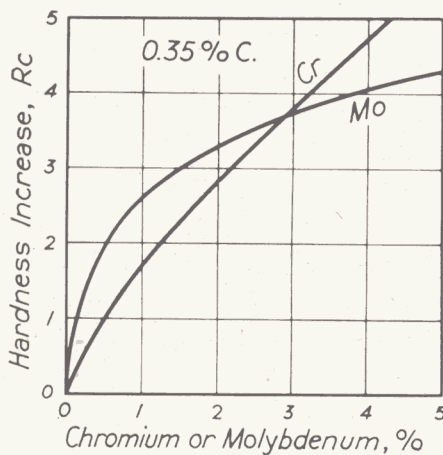


Fig. 87—Effect of Molybdenum and Chromium upon Maximum Martensitic Hardness in 0.35 Per Cent Carbon Steels. (See Footnote 23).

But here, again, some austenite is retained and it may be responsible for the lower hardness. The probability is that alloying elements do not lower pure martensitic hardness, other conditions being equal; rather their presence makes it improbable that pure martensitic hardness is realized.

Carbon and Martensitic Hardness — With reference to the effect of carbon upon martensitic hardness the effect is so marked and so consistent that one is forced to regard carbon as exceedingly fundamental in determining the properties of martensite. Small additions at low concentrations are relatively more effective than equal increments at higher con-

centration. Thus when the exceedingly high critical cooling rate for 0.02 carbon, low-metalloid steel is reached it is found that a Brinell hardness of 160 is achieved from this material of about 70 Brinell as annealed. This may appear to be a small degree of hardening, but it must be recalled that it is the same aggregate hardness brought about by 1 per cent carbon when the carbide is coarsely spheroidized in a commercial operation applied to carbon tool steel. In other words, this quench-hardening of 0.02 per cent carbon iron is equivalent to the aggregate effect of some 15 per cent of coarsely dispersed hard carbide. It has not been possible to reach the critical cooling rate in highly purified iron and accordingly the hardness of a martensitic pure iron is not known. Perhaps its hardness may be represented by an extrapolation of the maximum hardness curve for decreasing carbon content as shown in the top curve of Fig. 26. Actually, the most rapid quenches have not succeeded in producing even a very fine-grained ferrite in highly-purified iron.

The interest in the intrinsic hardness of quenched steel, or martensite, is, however, mainly academic. A very small proportion, indeed, of carbon or alloy steel ever reaches service in the untempered or martensitic state. Magnet steel may be left untempered and serve as permanent magnets in the as-quenched condition, wherein its mechanical properties do not come into play; a small amount of steel may be used in its hardest possible condition for glass cutters. Frictionless, ball or roller bearings, used at very high hardness, are generally somewhat tempered, and one may say that martensite is in the main only an intermediate product which must have its toughness increased, and in which little interest attaches to its precise hardness for the hardness will generally be reduced prior to use. To the capacity to achieve *approximate* martensite hardness for a given carbon content under increasingly slower rates of quenching we shall find much more reason to devote attention.

CONCEPTS OF HARDENABILITY

More than a passing reference has been made to two prominent circumstances in connection with the hardening of steel by quenching.

The first is that steels when hardened by martensitization may manifest only a certain maximum hardness which is dependent very largely upon carbon content, though the other components have a minor influence. That precisely this maximum possible hardness is not always realized is known, and is due to certain factors in the transformation but the minor departures are not of great importance. Quenched steels *approach* different intensities of hardening in accord with carbon content.

The second circumstance is that of the severity of quench, that is, the minimum rate of cooling, which will induce substantially martensitic hardness. A bar of one steel may harden completely in an oil quench while a similar bar of another would require the more rapid cooling of a drastic water quench and, even then, only the surface layers may escape transformation to the softer lamellar structure. Steels vary greatly in respect to the rate of quenching required to produce martensite and no characteristic of steels to be heat-treated is of greater importance, for this property determines also the maximum section which, in any given quenching technique, can be hardened through, or to any given depth below the surface. Since the milder quenches are in general less prone to cause cracking, the steels which do not require the severe quench are often requisite for larger sized parts.

It is important to distinguish critically between these two factors.

Intensity of Hardening versus Hardenability — In the instances wherein the hardness at the surface in the quenched and untempered steel is significant, a carbon content is gen-

erally selected which will fulfill the requirement and no complications arise, for only intrinsic hardness is involved. Suppose also that this martensitic hardness must be reached in the interior of a sizable section. Further increase in carbon content, while probably slightly augmenting the depth to which such hardening may be induced, will not be of any substantial avail. What is required is an enhanced capacity to develop the characteristic martensitic hardness for the steel under slower cooling rate. This capacity is practically independent of the intrinsic hardness of the martensite. Thus a fine-grained, high-carbon tool steel may easily reach the maximum hardness known in steels but only after the most vigorous quench in relatively small sections. Another low-carbon, deep-hardening alloy steel may achieve only a 58 Rockwell C hardness at maximum but at the same time this hardness will be found throughout the whole mass of a 3-inch or larger bar.

To this second characteristic we shall apply the word "hardenability" even though it would be a happy circumstance if some other word were available. To some users of steel, hardenability may, to some extent, call to mind the intensity of hardening when the steel is adequately quenched, but it is believed that the ability to achieve the martensitic hardness at minimum quenching rate will be the more logical meaning of the word. If only the two aspects of hardening were not in essence dissimilar one could combine the two in the concept of hardenability; instead, it seems necessary to emphasize the distinction.

Measurable Manifestations of Hardenability — Hardenability is the name here applied to the particular property of steels at elevated temperature (when so heated as to induce the austenitic state) which determines their ability to just transform to a substantially martensitic condition upon cooling at some minimum rate. This minimum cooling rate is the "critical cooling rate" and hence is, above all, the most

direct measure of hardenability. Thus if a 1-inch round of one steel fails to harden with an oil quench and requires the more rapid water quench, its hardenability is lower than that of a steel which will develop essentially martensitic hardness by quenching in oil. Hardenability is manifest also by the failure of large sized bars to harden throughout and, instead, in some certain quenching procedure to harden only in the outer zones. Thus the penetration (below the surface) of full hardening, or some certain degree of hardening, to greater or less depth in uniformly-sized and similarly quenched specimens becomes a measure of hardenability. This circumstance is inescapable since the interior of any cooling mass of steel must inevitably decrease in temperature at a slower rate than the exterior rate during the *early* stages of cooling, upon which alone the hardening of steel depends.

Another manifestation of hardenability is in the size of a piece which will just exactly harden through in a given quenching procedure. This limiting diameter of a suitable length of round then becomes a measure of hardenability. Grossmann²⁴ employs the diameter of round which will just be half martensitic at the center after a certain quench as a measure of hardenability and terms this diameter the critical size for that quench. In all tests supplying this type of information on hardenability, bars are employed of such a length that a section approximately midway between the ends is not subject to any end-effect.

In another test,²⁵ suited to inhomogeneous, e.g., carburized, sections, the quenching medium at some certain flow-velocity is applied only to one end of the bar while the other end cools nearly at the natural "air-cooled" rate. The distance from the quenched end at which some particular minimum hardness is reached then becomes an index of

²⁴"Hardenability as it Affects Heat Treated Parts", by M. A. Grossmann, *Metal Progress*, April 1938, p. 373.

²⁵"A Hardenability Test for Carburized Steel", by W. E. Jominy and A. L. Boegehold, *TRANSACTIONS, American Society for Metals*, Vol. 26, 1938, p. 574.

hardenability. Such tests are applicable to small bars of steel of fairly high hardenability.

These observations of hardenability, except the last, all require some knowledge of the hardness distribution across a transverse middle line or diameter of the quenched piece, or of the microstructures developed. Fig. 88 shows typical hardness distribution across the mean diameter of two bars of the same 0.9 per cent carbon steel. Certain shortcuts may be taken, however, to estimate hardenability according to some of the concepts. For example, the etching characteristics of hardened and unhardened steels are sufficiently different to make it possible to locate the transition zone in partially hardened sections with assurance. In fact it is not even entirely necessary to grind a smooth surface upon the section; a fracture surface is rapidly blackened by suitable etching to permit measuring the depth of the light etching hardened zone. Fig. 89 shows a group of 1-inch carbon steel rounds etched to reveal depth of hardening. This method is applicable to steels of low hardenability (carbon and low-alloy steels) but not so well suited to deeper hardening steels, because the transition zone is rendered too broad. In shallow hardening steels, however, the transition zone, composed of a mixture of martensite and fine pearlite, is narrow and the point naturally selected as the boundary for measurement is found consistently to be at the 50 per cent martensite circle.

A circumstance in connection with the transition zone of moderately shallow hardening steels makes the particular zone at 50 per cent martensite an excellent criterion of depth of hardening. It happens that the most rapid change in hardness occurs in the range between some 40 and 60 per cent martensite. This observation led Grossmann to employ the 50 per cent martensite zone as a basis of depth of hardened zone and for the critical size, even though a higher propor-

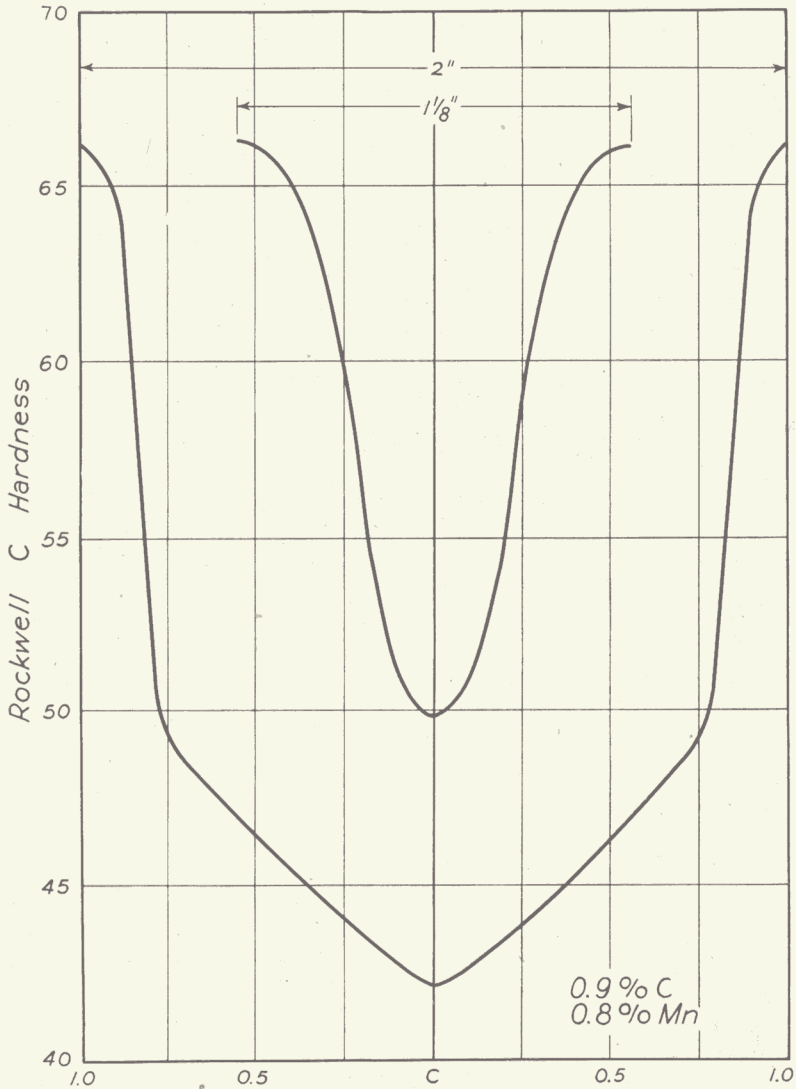


Fig. 88—Hardness Distribution across the Mean Diameter of a $1\frac{1}{8}$ -Inch and a 2-Inch Bar of the same 0.9 Per Cent Carbon, 0.8 Per Cent Manganese Steel Quenched in Water. (After Data of Grossmann, Asimow and Urban).

tion of martensite would have had a more practical significance.

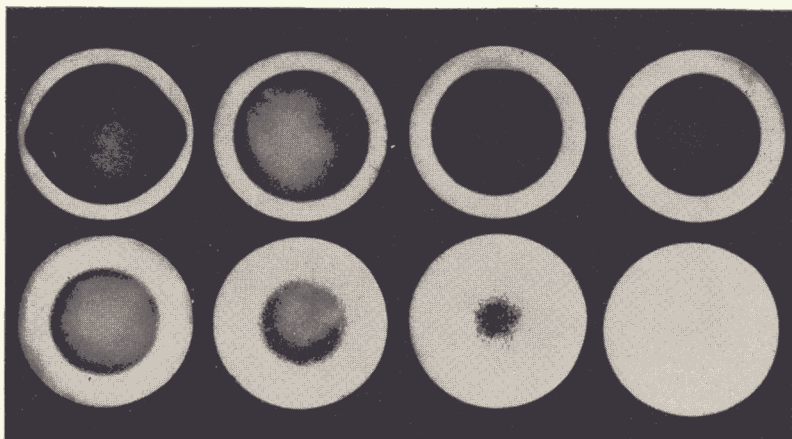


Fig. 89—A Group of Etched Mid-Sections from 1-Inch Quenched Bars of Carbon Steel.

Estimations and Designations of Hardenability — For some eight years the author and his associates have reported the complete distribution of hardness across suitably-sized, quenched rounds as a full description of hardenability, under a specific set of quenching conditions. A graph showing such data would appear to be an adequate representation of hardening behavior and, when shown for a series of sizes, comprises a fairly complete description of the hardenability of the steel under standard quenching conditions. That even this extended presentation of information has some minor, inherent shortcomings with respect to interpretation will appear later. Fig. 90 shows a number of these hardness plots; they correspond actually to the etched sections of Fig. 89, and refer to several 1 per cent carbon steels of substantially identical, ordinary chemical analysis but having different austenitic grain sizes as quenched. It is quite natural that a still simpler means of designating hardenability should be sought and, if possible, one which is directly numerical in character.

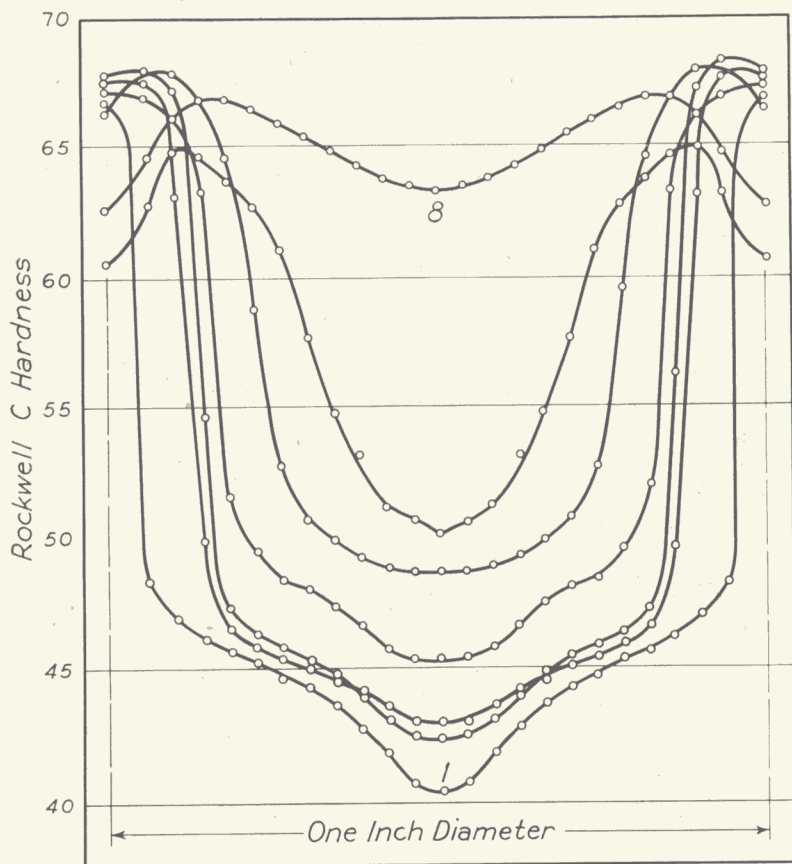


Fig. 90—Hardness Distribution Curves for the 1-Inch Diameter of the Steel shown in Fig. 89.

Burns and Archer²⁶ suggest securing the hardness distribution data, but designating hardenability by three numbers, first surface hardness, S, second, average hardness across a diameter (or radius), A, and third, center hardness, C. The "S-A-C" numbers are reported as Rockwell C hardness. For reasons which the proponents of this method of designation do not explain, they term the average hardness

²⁶"Quantitative Hardenability", by J. L. Burns, T. L. Moore and R. S. Archer, TRANSACTIONS, American Society for Metals, Vol. 26. 1938, p. 1.

across the diameter "The Area Under the Curve" and give it the dimension "Rockwell-Inches". This highly elaborate name is obviously not incorrect but would appear needlessly to complicate the designation. As applied to one type of steel at a time, the scheme has certain merits provided the hardness at sufficiently numerous concentric circles is known. Some, however, do not favor the confusion of hardness with depth of hardening, for by this scheme a high-carbon, shallow-hardening steel would bear the same A number as some lower carbon, deeper-hardening steel. Accordingly, when the designation is used the S and C numbers should not be omitted. For any groups of carbon steels of essentially the same range of composition the method is particularly useful in rapid comparisons. The S-A-C numbers do not, of course, describe the mode of distribution of hardness, and thus would fail to distinguish between a sharp and a more gradual transition zone. The "A" figure has been employed very effectively for shallow-hardening steels by Riegel and Burns²⁷ as measure of hardenability in showing the effect of certain elements upon hardenability.

The choice of a method of estimating depth of hardening and then the choice of a designation for it should be made upon the basis of the kind of steel under consideration, the dimensions available and particularly the nature of the application.

For carbon steels it frequently suffices to quench either a round bar or a disc and, after fracture, to etch in such a manner as to reveal the location of the usually sharp zone of transition from martensite to fine pearlite. It is generally not advisable to depend wholly upon the fracture texture to estimate depth of hardening; it may be misleading. At the same time the fracture of the fully hard zone may well be

²⁷"Hardenability of Plain Carbon Steels", by John L. Burns and Glen C. Riegel Symposium, Hardenability of Alloy Steels, American Society for Metals, 1939, p. 262.

examined for fracture grain size. A full description of this method is given by Shepherd.²⁸ Generally speaking, a more complete description of the hardenability is secured from a plot of the hardness across a representative diameter of the specimen, employing a round.

For steels of greater hardenability, particularly when the sections available are of a size to harden throughout when wholly quenched, the method of Boegehold and Jominy²⁹ is very advantageous. This end-quenching provides a greater range of cooling rates than are found in a small round of steel quenched all over. The removal of a layer of surface metal along the length of the bar (to provide a strip for hardness impressions) should be uniform, so that the hardness values presented thereon may be comparable. This method is almost the only one suitable for case-carburized steels; in this application particularly, the depth of metal removed must be rigidly uniform and not too great. The disadvantage of the method is the lack of sensitivity in shallow-hardening steels which harden only near the quenched end-surface and its supersensitivity in steels of great hardenability which harden nearly to the air-cooled end of the specimen as shown in Fig. 91.³⁰ Thus the relative difference in hardenability may be more as between two steels which differ only $\frac{1}{16}$ inch in distance at low hardenability than between two of higher hardenability which show a full inch's difference at the end remote from the quench.

For more or less routine verification of the hardenability of sizable sections of medium- and high-hardenability steels it often suffices to take only a few readings of surface hardness and a few readings near the center of a representative

²⁸"The P-F Characteristics of Steel", by B. F. Shepherd, TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 979.

²⁹See also "Hardenability Tests", by W. E. Jominy. Symposium, Hardenability of Alloy Steels, American Society for Metals, 1939, p. 66.

³⁰"Transverse Hardness Tests of Heat Treated Steels", by Gordon T. Williams. Symposium, Hardenability of Alloy Steels, American Society for Metals, 1939, p. 215-221-236.

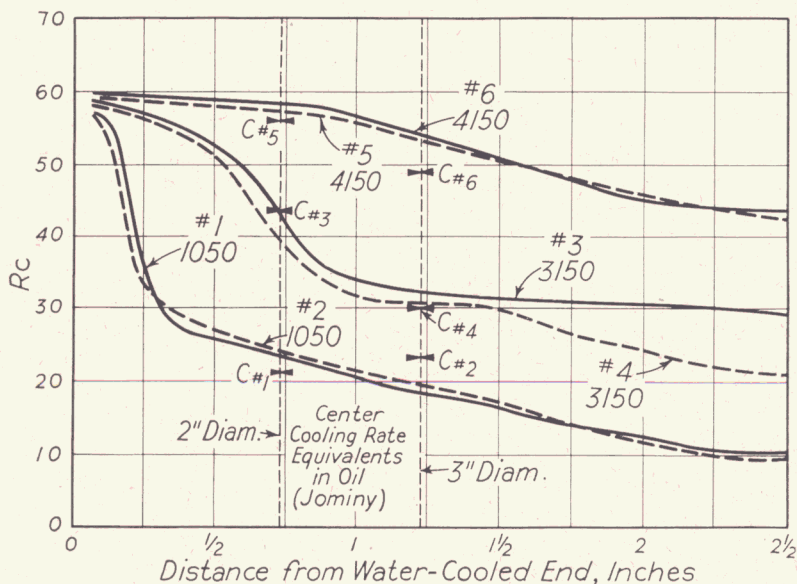


Fig. 91—Surface Hardness along the Specimens of Three Steels in the Boegehold-Jominy Test. (Williams).

section, particularly in the case of familiar steels. Both steel and hardening practice are combined in the results and both are indistinguishably reflected.

From the purely scientific standpoint, the determination of the actual rates of transformation at various significant constant temperatures might appear to be the ideal basic measure of hardenability since clearly these rates are responsible for the hardenability. Actually these determinations of fundamental rate of transformation do not lend themselves readily to a computation of hardenability in useful form. The reason for this conclusion is apparent when it is considered that the moment the steel cools below the temperature of austenite stability it is subject to a continuously changing rate of nucleation. An approximation to the critical cooling rate may doubtless be made by dividing the temperature range into zones and summing the nucleation at each level. The

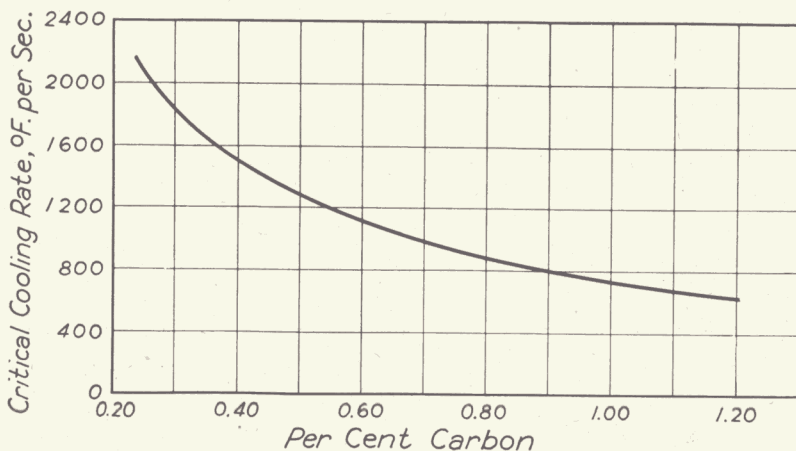


Fig. 92—Effect of Carbon Content upon the Critical Quenching Rate of Pure Iron-Carbon Austenite. (*Digges*).

direct determination of the critical cooling rate is probably easier to accomplish.

As an absolute measure of the hardenability of a steel, ready to be quenched, the critical cooling rate over an appropriate temperature range is probably the most definitive, but data are largely lacking for this approach to the subject. Digges³¹ has determined the critical cooling rate for iron-carbon alloys as shown in Fig. 92, and some data have been secured for manganese steels. Only one possibility suggests itself as perhaps invalidating this property as a practical and absolute measure of hardenability. It may well be, in principle at least, that the stress conditions alter velocity of transformation at the important temperatures and accordingly have an influence upon hardenability, so that perhaps the critical rates may not be precisely identical in different sized bars of the same material.

All in all, it would appear that a curve showing distribution of hardness across a radius or diameter of a quenched

³¹"Effect of Carbon on the Hardenability of High Purity Iron-Carbon Alloys", by T. G. Digges. TRANSACTIONS, American Society for Metals, Vol. 26, 1938, p. 408.

round bar constitutes a basis for becoming aware of the hardenability of a bar of heated steel quenched in any specific way. To make this description more fully qualify steel and quench, these plots may be shown for several diameters, although when the knowledge is required only for some certain size, one section is often sufficient. In the case of utilizing such information for other specific contours one should become familiar with the characteristic hardness-penetration of angles and special shapes for, as shown by Luerssen,³² the depth of hardness is greatly modified by the contour. It is pertinent to recall that the ratio of mass to heat dissipating surface in contact with the quenching medium varies as among large flats, cylinders and spheres.

The Physics of Hardenability — Probably the most exhaustive inquiry into hardenability and the physics of quenching steel is reported in a paper before the American Society for Metals by Grossmann, Asimow and Urban.³³ They evolved a philosophy of quenching and hardening steels, extending the concepts of the classical works of both H. J. French³⁴ and Howard Scott,³⁵ so that the extent of hardened zone in any round size of any steel may be computed for any evaluated quenching medium if only its behavior in any one medium (even unidentified) is well known. The assumptions for the system of Grossmann, Asimow and Urban are as follows:

1. Heat-transfer in accord with the law of Newton.
2. Equal heat-transfer constant for surfaces of low or high curvature, i.e., large or small sizes when similarly heated.

³²"Hardenability in Light Sections", by G. V. Luerssen. Symposium, Hardenability of Alloy Steels, American Society for Metals, 1939, p. 302.

³³"Hardenability, Its Relation to Quenching and Some Quantitative Data", by M. A. Grossmann, M. Asimow and S. F. Urban. Symposium, Hardenability of Alloy Steels, American Society for Metals, 1939, p. 124.

³⁴"A Study of the Quenching of Steels", by H. J. French, TRANSACTIONS, American Society for Metals, Vol. 17, 1930, p. 798.

³⁵"Some Problems of Quenching Steel Cylinders", by Howard Scott, TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 68. "The Problem of Quenching Media for the Hardening of Steel", by Howard Scott, TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 577.

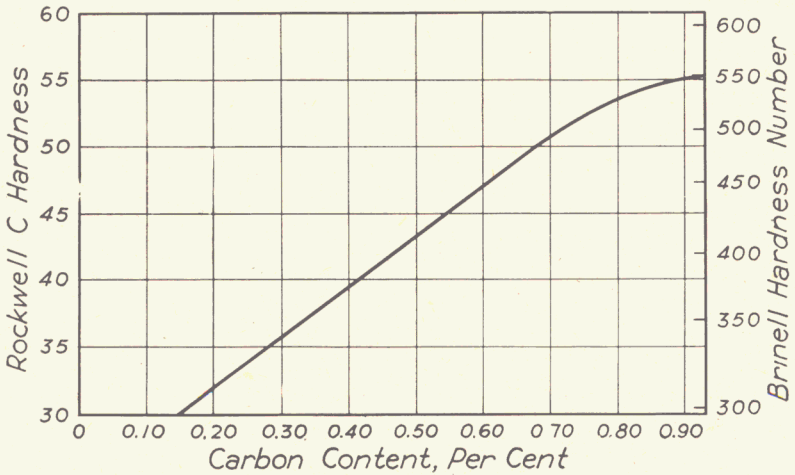


Fig. 93—Hardness Values for the Half-Martensite Condition in Quenched Carbon Steels of Various Carbon Contents. (Private Communication from M. A. Grossmann).

3. Effectively uniform and homogeneous austenite, center to surface, in large and small specimens of the same steel as heated.

With this set of assumptions, one may calculate the rel-

sizes of bars and for different quenches. Ascertainment of severity of quench is then based on the following simple axiom: assuming the same steel in various bar sizes, then all positions exhibiting the same hardness *after* the quench must have cooled alike *during* the quench. (As an example, after quenching a 0.50 per cent carbon steel in water, one may find a hardness 48 Rockwell C in a 1-inch bar at the center, and in a 1¼-inch bar at ¼ inch below the surface. One concludes that the cooling was the same at these respective positions in the two bars.) It turns out from the calculations that the knowledge of these positions enables one to ascertain the severity of the quench that was employed.

Hardenability tests are usually made by quenching, and

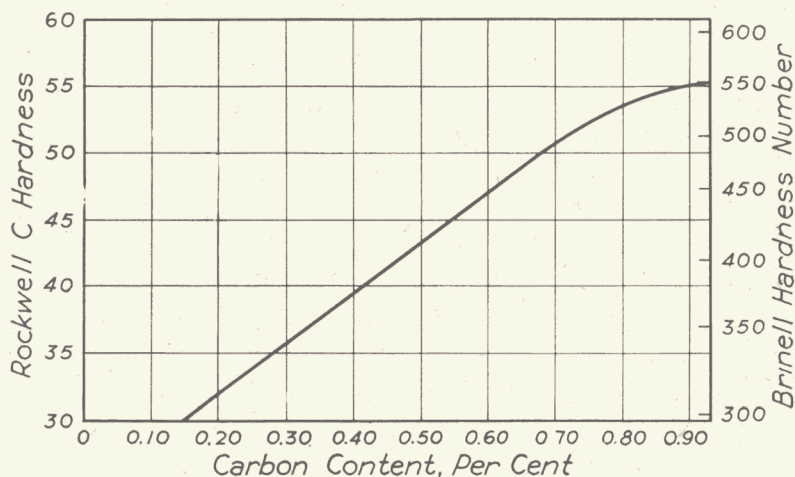


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3. Effectively uniform and homogeneous austenite, center to surface, in large and small specimens of the same steel as heated.

With this set of assumptions, one may calculate the relative cooling rates (or cooling times) at different positions (depths) within a bar that is being quenched, for different sizes of bars and for different quenches. Ascertainment of severity of quench is then based on the following simple axiom: assuming the same steel in various bar sizes, then all positions exhibiting the same hardness *after* the quench must have cooled alike *during* the quench. (As an example, after quenching a 0.50 per cent carbon steel in water, one may find a hardness 48 Rockwell C in a 1-inch bar at the center, and in a 1¼-inch bar at ¼ inch below the surface. One concludes that the cooling was the same at these respective positions in the two bars.) It turns out from the calculations that the knowledge of these positions enables one to ascertain the severity of the quench that was employed.

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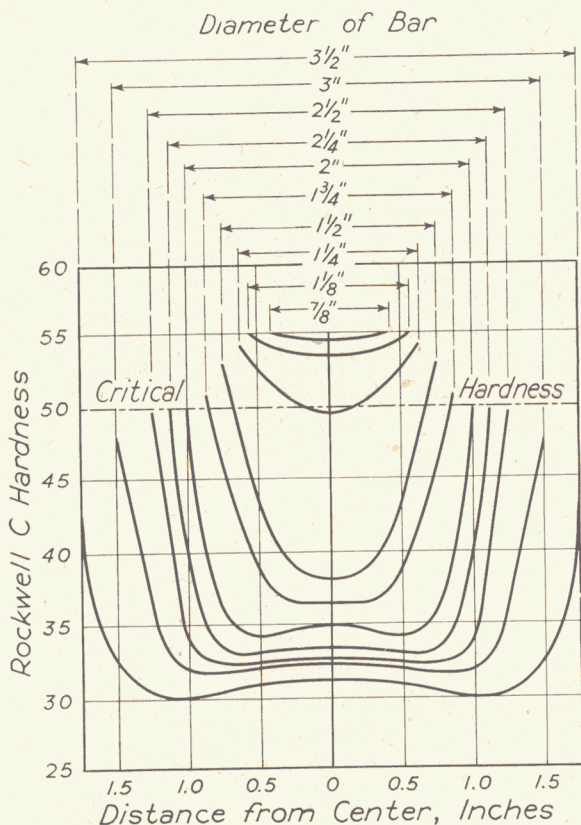


Fig. 94 (Part 1)—Hardness Distribution Curves for S.A.E. 3140 Steel Quenched from 1500 Degrees Fahr. into Oil. (Grossmann, Asimow and Urban).

since the hardness distribution in a quenched bar depends both on the hardenability of the steel and on the severity of the quench, the above-mentioned knowledge of the severity of the quench makes it possible to evaluate the *hardenability alone*. Their principal measure of hardenability is the "critical size" (diameter of cylinder) which will just harden through in the sense of forming just 50 per cent martensite at the center; this proportion is the criterion of depth of hardening also. This measure of hardenability was employed

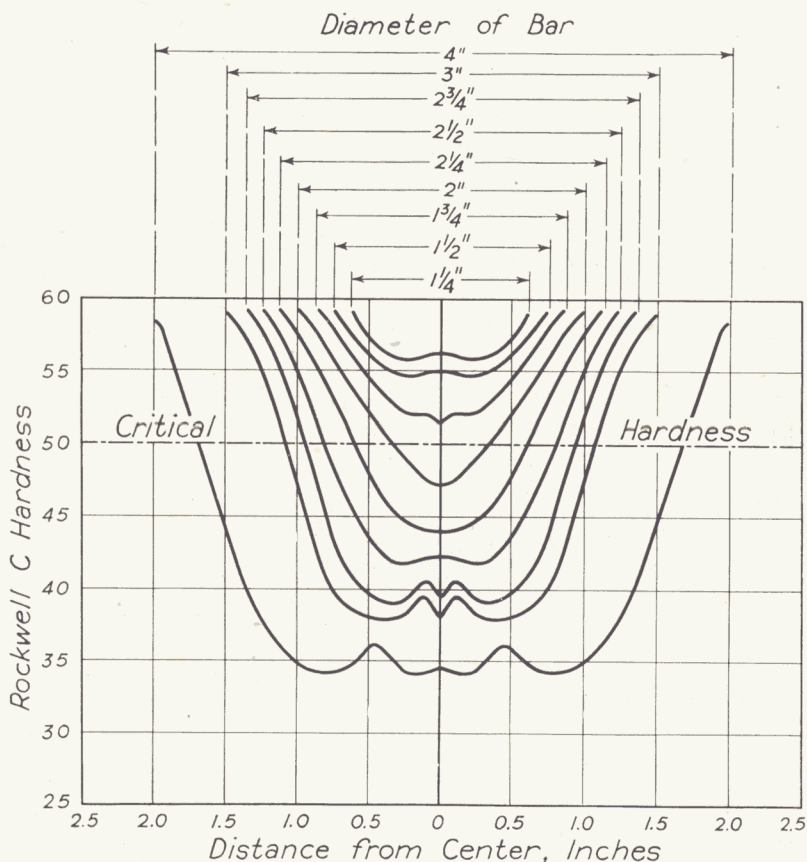


Fig. 94 (Part 2)—Hardness Distribution Curves for S.A.E. 3140 Steel Quenched from 1500 Degrees Fahr. into Water. (Grossmann, Asimow and Urban).

because the *position* of this structure in the bar can be determined with more accuracy than can any other proportion of martensite. The approximate value of the half-martensite mixture for different carbon contents is shown in Fig. 93. In the exemplification and verification of their computations they employed specially precise measurements of hardness distribution as indicated in their charts reproduced in Fig. 94. These exceedingly reliable data are consolidated in Fig.

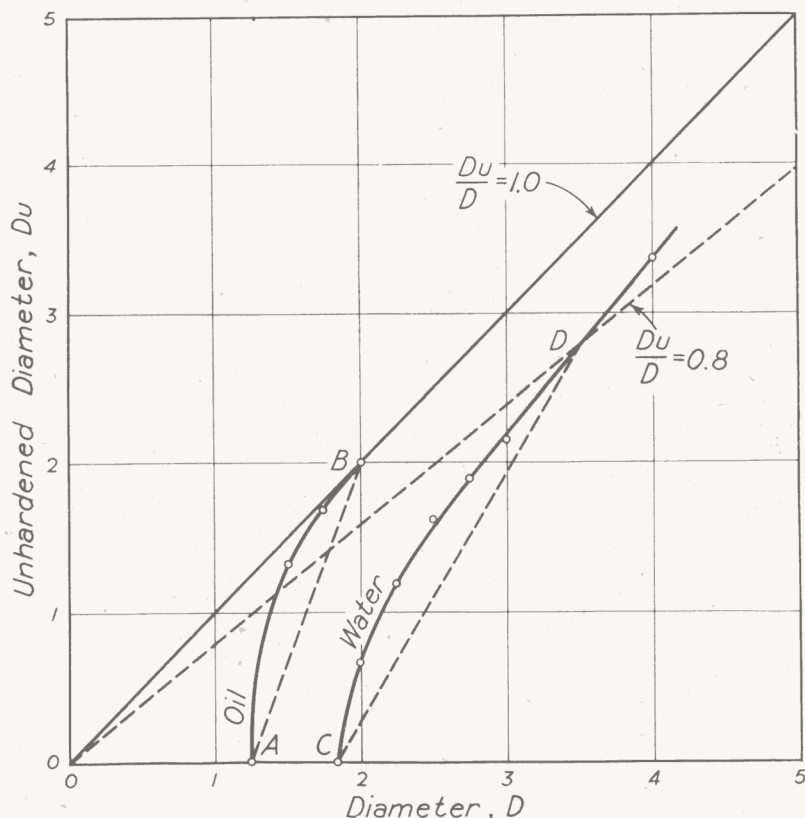


Fig. 95—Characteristic Curves, Unhardened Core Diameter versus Diameter, for the Quenches of Fig. 94. (Grossmann, Asimow and Urban).

95, in which the diameter of unhardened core, D_u , is plotted against diameter of bar D , both for a medium water quench and for a medium oil quench.

Their computations showed that any two steels, A and B, with respectively high and low hardenability could have the same critical size, D_o , ($D_u = 0$) as differently quenched, e.g., A in oil, B in water, but that their respective depths of hardening (A quenched in oil, B in water) at all other equal sizes for comparison would be different; that is, their D_u vs. D or

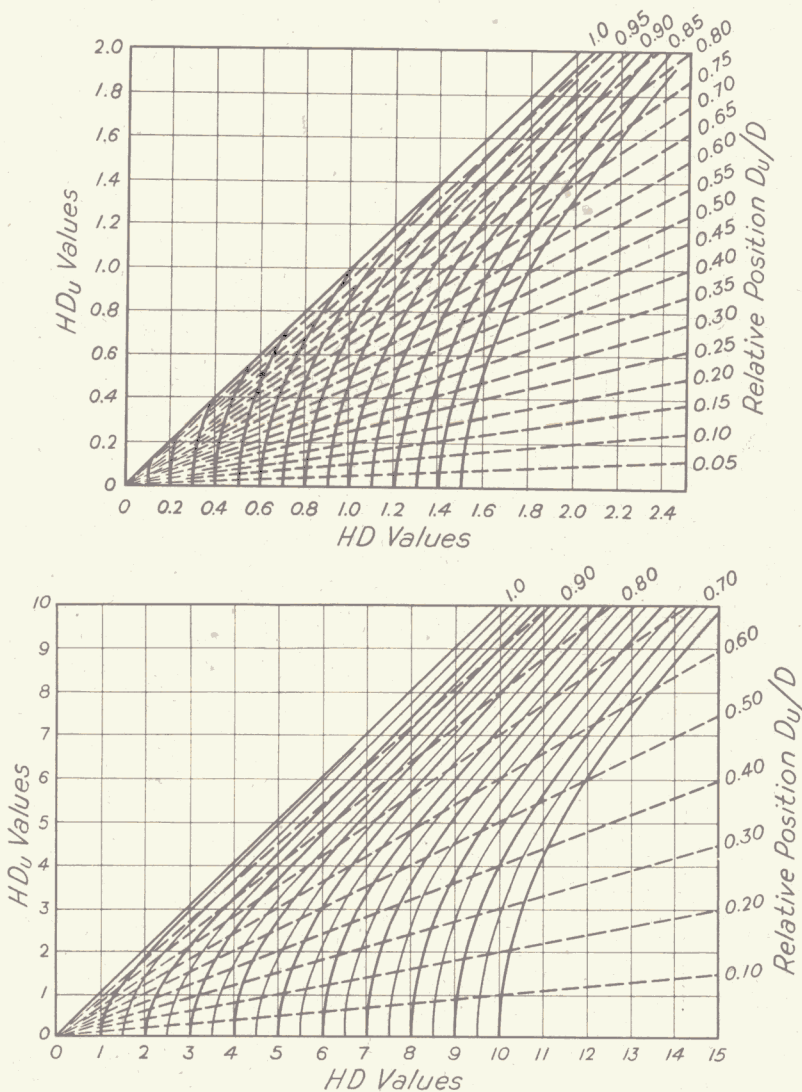


Fig. 96*—Detailed Characteristic Curves, HD_u versus HD , for Two Ranges of HD Values. (Grossmann, Asimow and Urban).

“characteristic” curves would always differ. They found, also, that the *form* of these “characteristic” curves was identical for a series of specimens and quenches *under one*

set of circumstances, covering the range from high velocity of quench for small sizes of steels of low hardenability up to large specimens of mildly-quenched, deep hardening steels. The identity in form existed when the product of the heat transfer factor, H , and the critical diameter, D , was a constant. So long as HD is constant all Du/D curves are the same if the scale of plotting is rightly chosen. Thus one characteristic curve represents an entire family when HDu is plotted against HD instead of Du vs. D . All steels and all quenches may be covered then in a family of HDu vs. HD curves (Detailed Characteristic Curves) as in Fig. 96. Constant values of Du/D are shown as dashed lines appropriately marked.

Thus if one has a characteristic curve such as those in Fig. 95, it is only necessary to match its *form* to one of the curves of Fig. 96, to find the HD_0 and hence the H value. Actually two short methods are introduced for this purpose involving the inclination at certain $\frac{Du}{D}$ values so that finally

the knowledge of the depth of hardening in two suitable sizes serves to identify in a unique manner both the hardenability of the steel (D_0) and the severity of quench (i.e., heat-transfer factor, H). This can be easily accomplished by the second shortcut which is the logarithmic chart of Fig. 97. Another chart enables one to ascertain the critical size under conditions of the ideal quench or any other quench of identified " H ". Fig. 98 shows the variation of critical size with the H value of a wide variety of quenching mediums, including the "ideal" quench which implies a medium of infinite H value, capable of reducing the surface of the steel to bath temperature in zero time interval.

The charts and methods of Grossmann, *et al.*, make it possible, moreover, to derive even the distribution of hardness in any size round after any identified quench provided

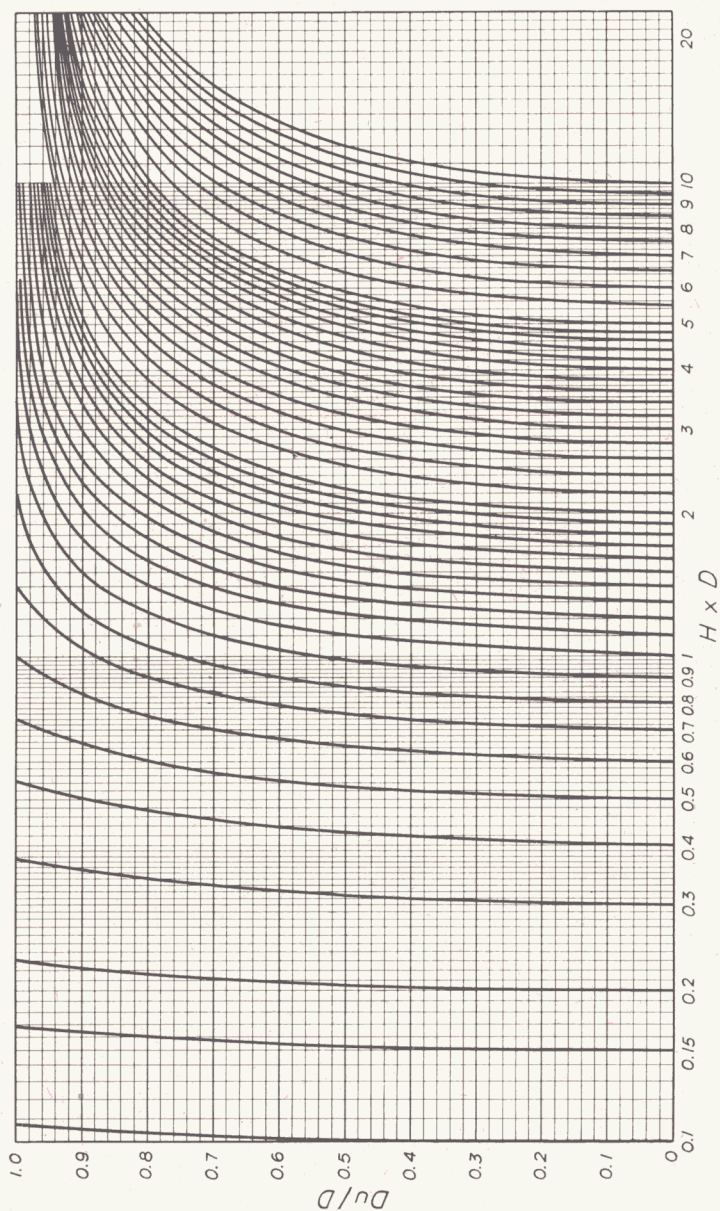


Fig. 97—Chart for Determining HD and hence H , with HD Plotted on a Log Scale versus D_u/D . (Asimow).

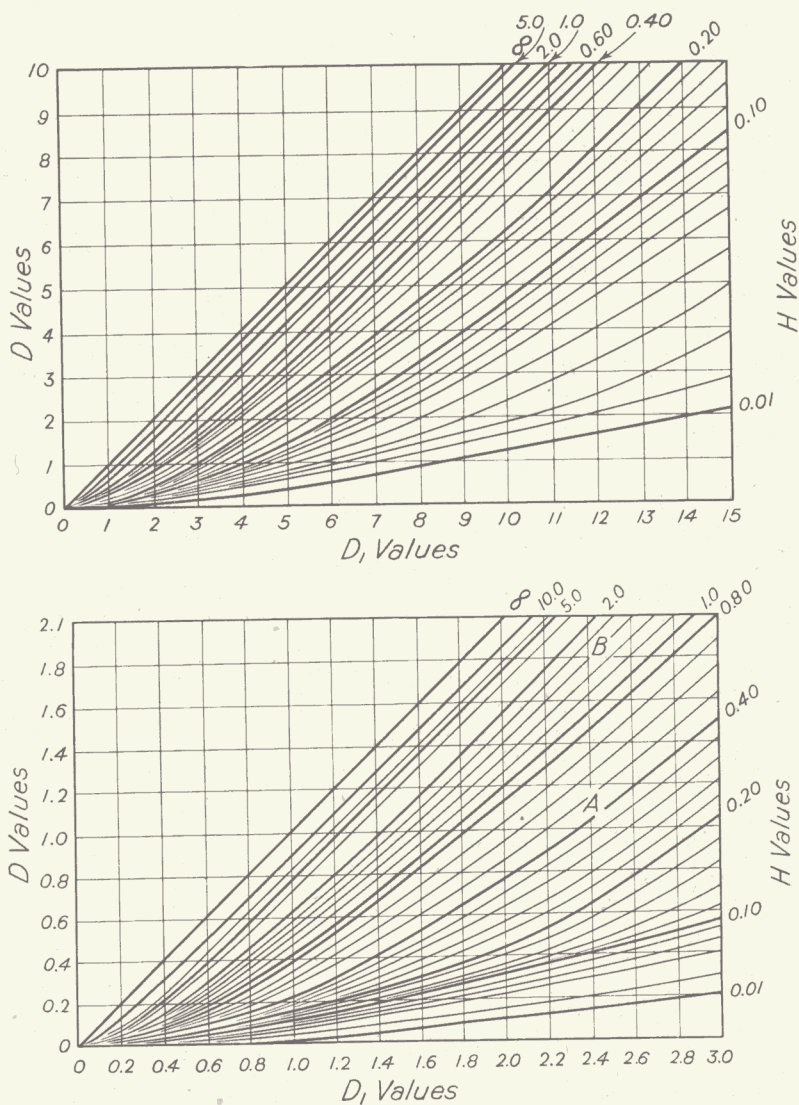


Fig. 98—Relationships among Ideal Critical Size D_i , Actual Critical Size D , and Severity of Quench H , Two Scales. (Grossmann, Asimow and Urban).

only its behavior in one suitable medium is well known. The discrepancy from actual performance may be at a maximum

in this particular operation but it is not in general so great as to render the prediction of no use; the other more useful computations are, in general, of useful accuracy.

The assumptions of the computations are probably not rigidly fulfilled in any respect in the quenching of different steels in different sizes but the "errors", that is, discrepancies, are in many instances amazingly small. The fact is that the concept makes clear such heretofore perplexing circumstances as (1) the persistence of a shallow-hardened zone in exceedingly large rounds, even of moderately low hardenability when the quench is very severe and (2) the little change of its thickness ($D - D_u$) with change in diameter. Similarly the abrupt disappearance of a hardened rim at fairly small size in steels quenched in mild mediums is clarified. In turn these confirmations show how soft spots are almost inevitable under certain conditions. It should be remembered that surface conditions of the steel which influence heat abstraction (e.g., scale thickness and texture) are comprehended in the factor "H" as well as thermal diffusivities. One may easily employ the two curves of Fig. 37 (with a hardness of about 53 Rockwell C at the 50 per cent martensite zone) to try out the Grossmann analysis. First employ curve of Fig. 97 to secure HD and H, and then discover D_I from Fig. 98. H will be found to be about 2.3; the critical size for this quench is 0.86 inch with a D_I of about 1.22 inch.

As a matter of interest the values of "H" for commonly applied quenches cover a wide range; they vary considerably more within commonly designated technique than would probably be expected. Thus the following are some approximate observed values:

<u>Mode of Quench</u>	<u>H Value</u>	<u>Mode of Quench</u>	<u>H Value</u>
Cooling in Still Air	0.02	Moderate Motion in	
Moderate Motion in Oil	0.4-0.5	Warm Brine	2.0
No Motion in Water	1.0	Moderate Motion in	
High Velocity Oil		Water	3.0-4.0
		High Velocity Water	5.0 and up

Finally it should be stressed that the analyses of Grossmann, Asimow and Urban lead them to report the hardenability of a steel as an ideal diameter which would just be rendered half martensitic at the center if quenched in a medium of infinite heat abstracting power. Thus the hardenability is reported as a characteristic wholly of the steel without any reference to any actual quench. Clearly this is an advantage not possessed by other schemes.

It should be remembered that the method depends upon associating a certain hardness and its zonal location with a certain rate of cooling. This being the case the philosophy applies in large part to any shape which may be quenched, even though the graphic tools of the analysis are prepared for circular cylinders. With appropriate minor modifications the entire scheme may be adapted to unilateral quenches as in the Jominy test and it is to be hoped that the authors will develop the scheme for still wider application with the same ingenious kind of graphic solutions possible.

Queneau and Mayo³⁶ have proposed an approximate method for determining the critical size from two specimens (different diameter) of a steel as identically heated and quenched. They plot the ratio, $\frac{h}{r}$, of hardened rim thickness, h , to radius of round, r , against the inverse square of the diameter, $\frac{1}{D^2}$. With two points representing two diameters they draw a straight line prolonged to the value $\frac{h}{r} = 1.0$. This intersection marks the approximate critical size. Actually the assumption that $a \frac{h}{r} = \frac{1}{D^2} + b$ holds only for the ideal quench. So long as the quench is vigorous the approximation is reasonably close. The treatment of data is shown

³⁶"Hardenability and Its Designation, the Hardenability Line", by B. R. Queneau and W. H. Mayo. Symposium, Hardenability of Alloy Steels, American Society for Metals. 1939, p. 237.

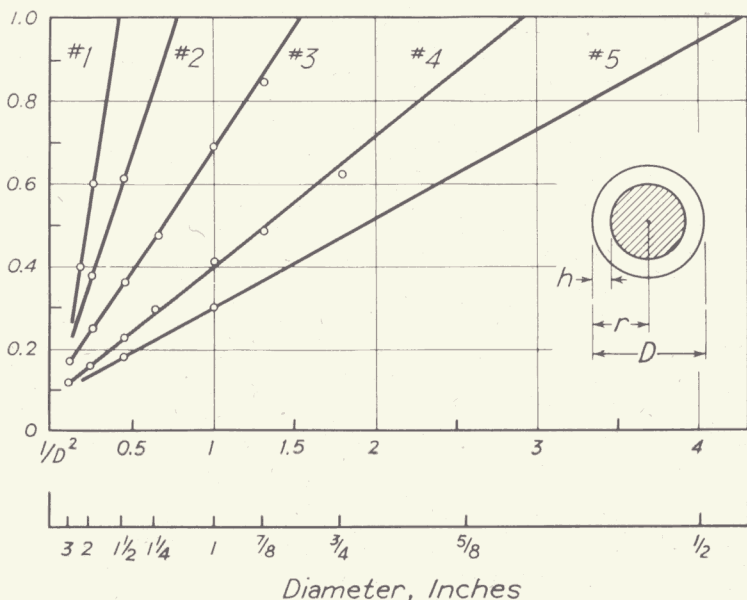


Fig. 99—Proportion of Hardened Exterior Zone Plotted against the Inverse Square of the Diameter of Quenched Rounds, according to Queneau and Mayo.

in Fig. 99. This straight line is called the “hardenability line” and its slope is a useful indication of the relative hardenability of the steel under the conditions of a given quench. Obviously the discrepancy between fact and assumption shows up as an error in the estimate of hardenability.

Summary of Hardenability Designations—A few broad generalizations may be mentioned even in a field of study which is currently being developed rapidly.

Where sections of sufficient size to develop only incomplete hardening at the center are available, probably the most informative measure of hardenability is the distribution of hardness across the mean diameter of rounds. Whenever possible two sizes differing enough to make an emphatic disparity in hardness penetration should be explored. The analysis of the characteristics of the quenching practice and

the steel may then be made by the system of Grossmann and his collaborators.

This analysis can also be carried out in steels of not too great hardenability by using the depth of hardened zone revealed by suitable etching. In measuring hardness distribution the impressions are preferably made not upon actual diameters but upon concentric circles in such a way as to "stagger" the readings which should be numerous. A fixture in the form of the horizontal V attached to the hardness tester will insure concentric readings. It can be advanced by a screw which insures uniform radial spacing.

The analysis of Grossmann will reveal how easily inequalities in successive quenches may be set up in poorly controlled practice; two adjacent sections from the same steel bar are about as nearly identical as two separate entities can possibly be.

So long as steel is cast into ingot molds there will be certain systematic variations of carbon (and manganese and phosphorus) as between center and surface. This circumstance is reflected in anomalies in the hardness distribution curves and possibly even in determination of the critical size. Most heat-treated parts are loaded in such a manner as to put relatively little stress upon the center and the circumstance of ingot segregation is of limited practical significance.

Steels of relatively high hardenability for which limited sections only are available are advantageously rated for relative hardenability by the single end quench according to Jominy. Clearly the test should be used for carburized steel wherein the true hardenability of the carburized zone is of interest.

Only with an adequate basis of comparison is it possible to consider the contribution to hardenability of the various elements. For this purpose we shall mainly employ the hardness distribution curve for round bars.

EFFECTS OF ALLOYING ELEMENTS UPON HARDENABILITY

In the course of gaining some familiarity with the interrelations between hardenability of steel and the severity of the quench employed in hardening, it has become apparent that some of the alloying elements greatly extend the range of sizes which may be substantially hardened by any certain quenching procedure. Thus, for example, a vigorous water quench will result in 50 per cent martensite at the center of a round of S.A.E. 1045 steel about 1 inch in diameter (the critical size for $H = 4$ to 5), while an equally vigorous quench will bring about a similar state of structure in a round of S.A.E. 3140 nearly 2 inches in diameter. The presence of some 1.25 per cent nickel and around 0.60 per cent chromium just about doubles the hardenability of the steel in terms of critical size, as may be seen at a glance from Figs. 38 and 95.

Now, needs exist for steels of like carbon content but with far greater hardenability,—not only with water-quenching practice,—but, in the interest of reducing quenching stress, securable even with the milder oil quench. The use of alloying elements supplies the need and at the same time a suitable choice of the particular deep-hardening elements may often result in securing in addition other enhanced properties.

It would be interesting here to speculate upon the vast differences from the existing modern machine design which would have been necessitated had the alloying elements not retarded the austenite-pearlite reaction and thereby contributed hardenability. Time does not permit an extended excursion into such fancy, but one may visualize layered members, in tension particularly, being built up of thin heat-treated sections and of large members, limited to modest working stress, having only the toughness of lamellar structures in carbon steel. Probably engineers would have managed, but

at a cost; appearance might have been quite different for many machines.

Factors Influencing Hardenability—Now, hardenability is altered, as has been shown in Fig. 36 by factors other than composition of the austenite solid-solution itself. In a single composition the hardenability was shown to be a function of the austenitic grain size existing at the moment of quenching. The grain size has such an effect because, apparently, it exercises a control over nucleation capacity. The majority of nuclei for the A_r' transformation, which must be escaped in hardening, are formed in the austenitic grain boundaries and it is logical to expect that fine grains, supplying vastly greater grain-boundary surface, would likewise increase nucleation and reduce hardenability.

Quite clearly, also, foreign particles of some types contrive to stimulate (or themselves serve as) nucleation centers for the pearlitic transformation. Hence not only grain size, but inclusions which may also restrict grain growth, act *directly* as well to influence hardenability. Experimental evidence is not lacking to confirm the double role of certain fine dispersions of particles to promote nucleation directly and to act as inhibitors of grain growth, — both effects conspiring to reduce hardenability. When we speak of elements *dissolved* in austenite as influencing hardenability we shall of necessity have to be at pains to compare steels with similar grain size and inclusions at the time of quenching. When such properly restricted comparisons are made it is found that dissolved elements, with the exception of cobalt, all contribute to hardenability but that they have quite different degrees of influence, small amounts of some being as effective as much larger amounts of others.

Another influence upon hardenability which is not characteristic alone of any certain elements is exerted by the degree of homogeneity of the austenite. When the austenite

of some certain mean composition is richer in some zones than in others with respect to alloy, the hardenability is lowered in the sense that the lean zones transform to fine pearlite at high quenching velocity and a pure martensite is difficult to secure. On the other hand, the alloy-rich regions transform slowly and would become martensitic even at low quenching rate. Elements which segregate to form special carbide and then diffuse slowly oppose homogeneity.

It would be possible to speak only of the over-all effect of an element upon hardenability of steel under a prescribed set of heating conditions, ignoring the mode of its distribution, — that is, giving no heed to whether the element was wholly dissolved in austenite or in part incorporated in carbide or inclusions. But this empirical approach would be uninteresting and unfruitful, because certain anomalies would remain unexplained. It is more enlightening to adopt an analytical viewpoint and consider the effects of the elements upon hardenability in the light of their mode of distribution. For this method of analyzing the problem let us enumerate the factors which fundamentally influence hardenability.

A. Decrease Hardenability —

Factors which hasten nucleation:

1. Fine Grains of Austenite.
2. Undissolved Inclusions.
 - a. Carbides (or Nitrides).
 - b. Nonmetallic Inclusions.
3. Inhomogeneity of Austenite.

B. Increase Hardenability—

Factors which retard nucleation or growth of Ar' products from nuclei:

1. Dissolved Elements in Austenite (except Cobalt)
2. Coarse Grains of Austenite.
3. Homogeneity of Austenite.

Consolidating these items without respect to the direction of the influence there are then 5 principal (but mutually interrelated) factors:

1. Mean Composition of the Austenite.
2. Homogeneity of the Austenite.
3. Grain Size of the Austenite.
4. Nonmetallic Inclusions in the Austenite.
5. Undissolved Carbides (and nitrides) in the Austenite.

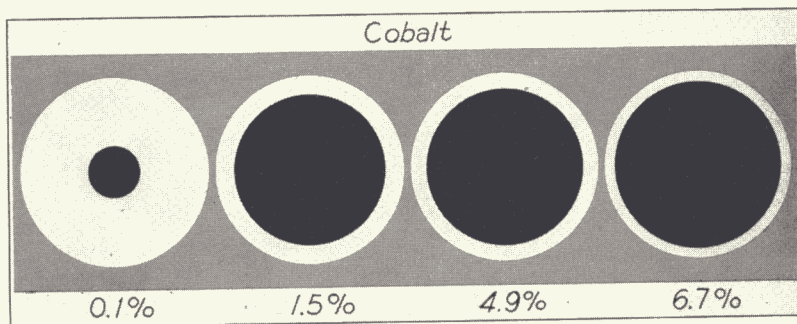


Fig. 100—Depth of Hardening (White Circumferential Zone) in a Series of 0.90 Per Cent Carbon Steels of Increasing Cobalt Content as Indicated. All Quenched from 1740 Degrees Fahr. (*Data of Houdremont*).

R. F. Mehl³⁷ has contributed a brilliant analysis of the probable working of these factors and his study of the dissimilarities of these major effects is well worth careful consideration.

It will be appropriate to consider these factors as they are made more or less effective by the several alloying elements.

Elements in Solution in Austenite—How cobalt, of all the elements which dissolve in austenite (and all do so to a substantial extent), should lower hardenability is unknown. It is extraordinary to find that it alone hastens the trans-

³⁷"The Physics of Hardenability", by Robert F. Mehl. Symposium, Hardenability of Alloy Steels, American Society for Metals, 1938, p. 1.

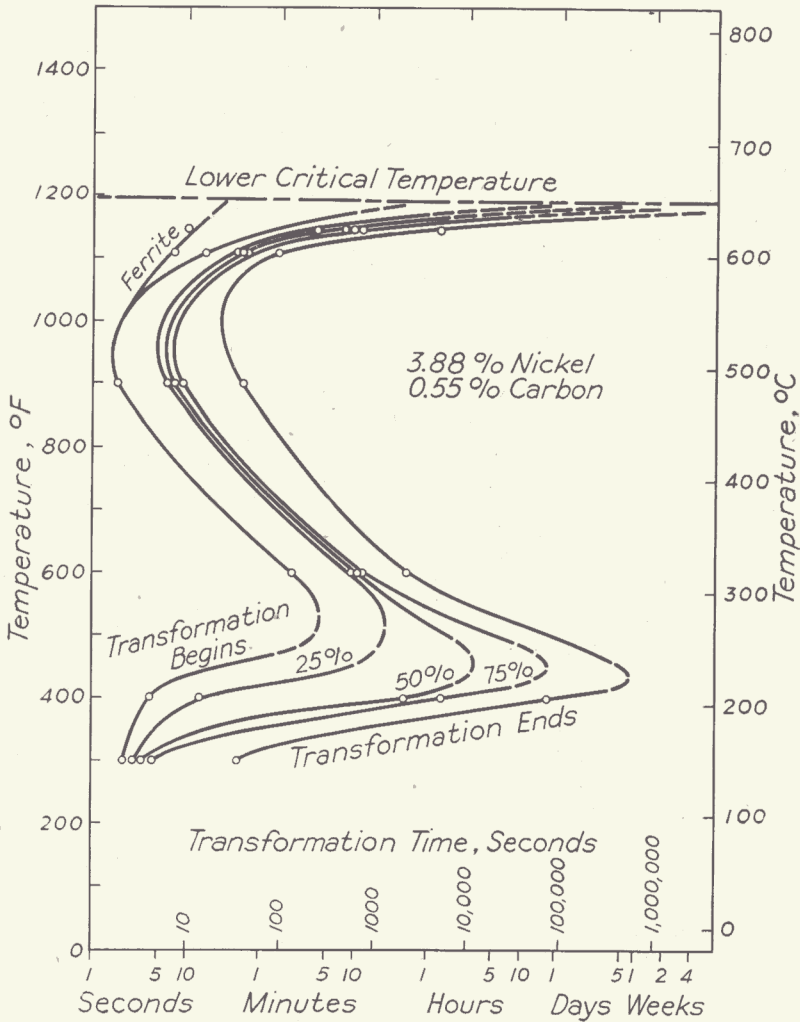


Fig. 101—The S-Curve of Transformation Rates for a 0.55 Per Cent Carbon, 3.9 Per Cent Nickel Steel. Heated to 1600 Degrees Fahr. (870 Degrees Cent.).

formation to the softer products at Ar' rather than retards it, and that in its presence hardening may occur only with increased cooling velocity. Fig. 100 shows the depth of hardened rim, as found by Houdremont, in a series of 0.90

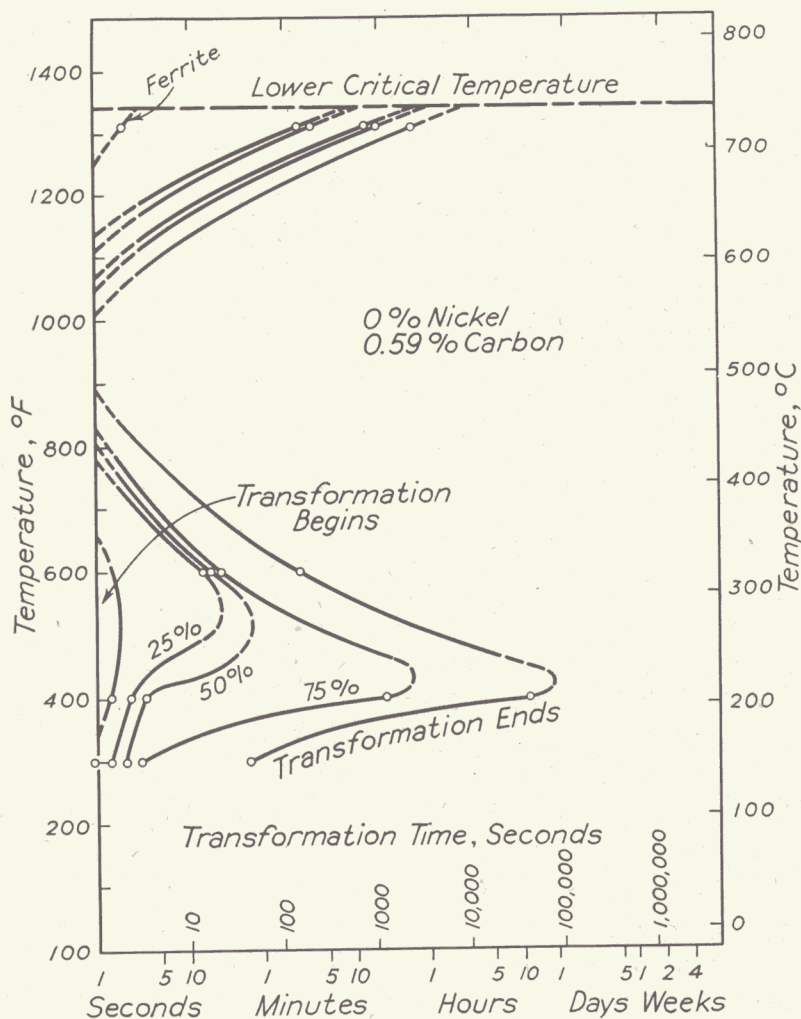


Fig. 102—The S-Curve of Transformation Rates for a 0.55 Per Cent Carbon Steel Comparable with that of Fig. 101 except for Nickel. Time Intervals on Logarithmic Scale.

per cent carbon steels with increasing cobalt content all as quenched from 1740 degrees Fahr. (950 degrees Cent.). The circumstance has been verified by the author and others who have failed to induce martensitic hardness in a 1-inch round

of 10 per cent cobalt steel although a transformation certainly occurred. Having noted this amazing exception to the rule it may be dismissed as being at present unexplained.

Turning to the other elements, which retard transformation and induce hardenability, it is of some interest to verify the fact of increased reaction time at constant temperature induced by dissolved elements. Fig. 101 in comparison with

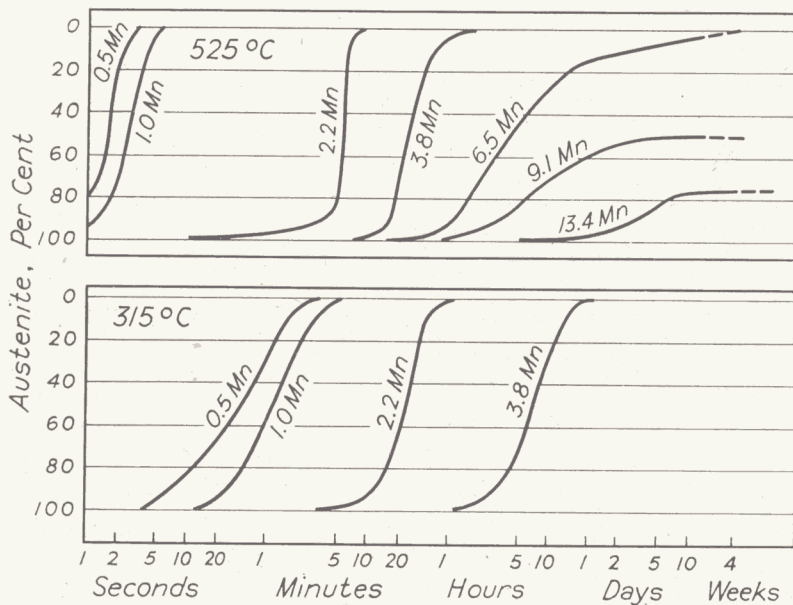


Fig. 103—The Individual Transformation Curves, from 100 Per Cent Austenite, at 980 Degrees Fahr. and 595 Degrees Fahr. in a Series of Manganese Steels of about 0.55 Per Cent Carbon. Time Intervals on Logarithmic Scale.

Fig. 102 well illustrates the retardation of transformation at all temperatures brought about by about 4 per cent nickel. Similar retardations are brought about by other elements in the significant temperature range above about 1000 degrees Fahr. (540 degrees Cent.). The even greater retardation brought about by manganese at a temperature of 980 degrees Fahr. (525 degrees Cent.) is shown in Fig. 103. The upper

curve for 2.2 per cent manganese has an early beginning of transformation but the bulk of the austenite transforms only after considerable time. This condition strongly suggests a lack of homogeneity in the austenite of the particular specimen employed,—a condition not improbable in manganese steel of this composition.

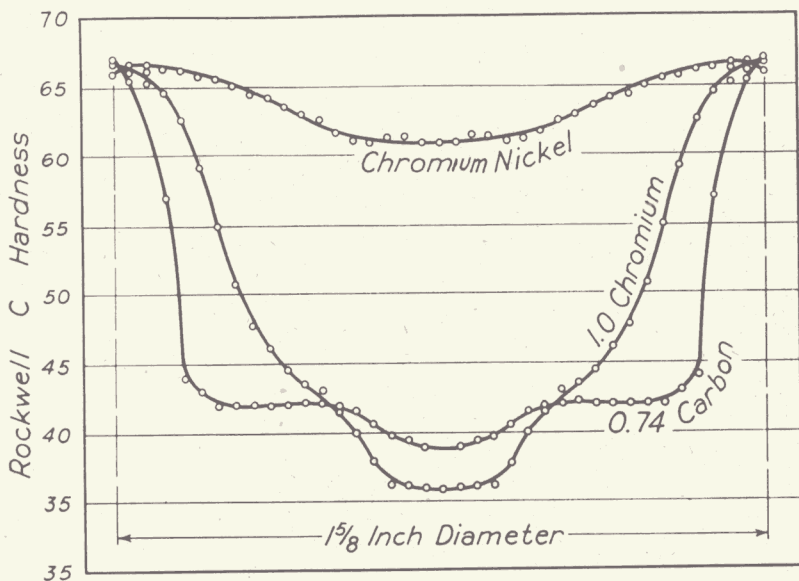


Fig. 104—The Distribution of Hardness across the $1\frac{5}{8}$ -Inch Diameters of Three 0.75 Per Cent Carbon Steels. Effect of Alloy may be Inferred by Comparison with Lower Curve for Carbon Steel.

The actual practical increase in hardenability brought about by dissolved elements is illustrated in the three hardness distribution curves of Fig. 104 for $1\frac{5}{8}$ -inch rounds of about 0.75 per cent carbon steels, two of which carry substantial amounts of alloy. The warm brine quench applied here had a heat abstracting power of $H = 2.3$. Thus about 1 per cent chromium increases the critical size for this quench from 0.86 inch to 1.2 inch or the Ideal Critical Size (D_I) from 1.23 to about 1.65 inch. This comparison is not precise because other elements were not absolutely constant.

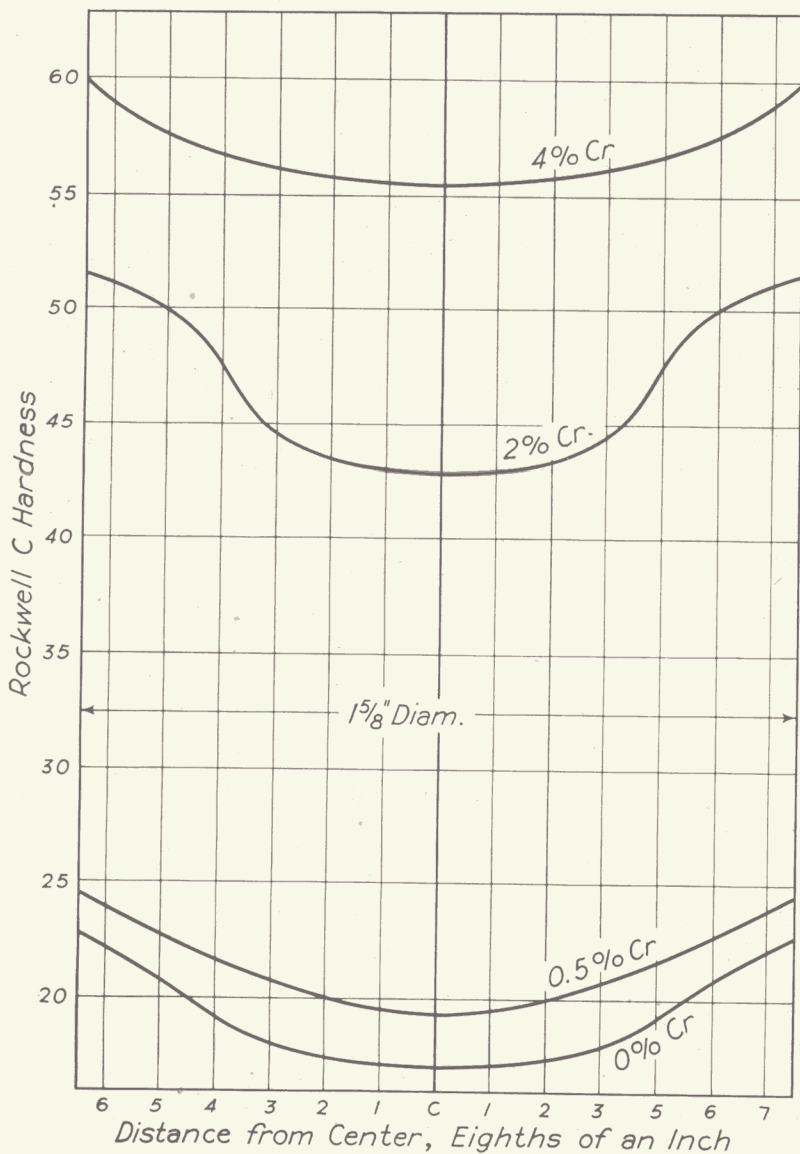


Fig. 105—The Distribution of Hardness across 1 5/8-Inch Rounds of Four 0.35 Per Cent Carbon Steels with Chromium as Indicated. The Bars as Quenched in Oil were Free from Undissolved Carbide. (See Footnote 38).

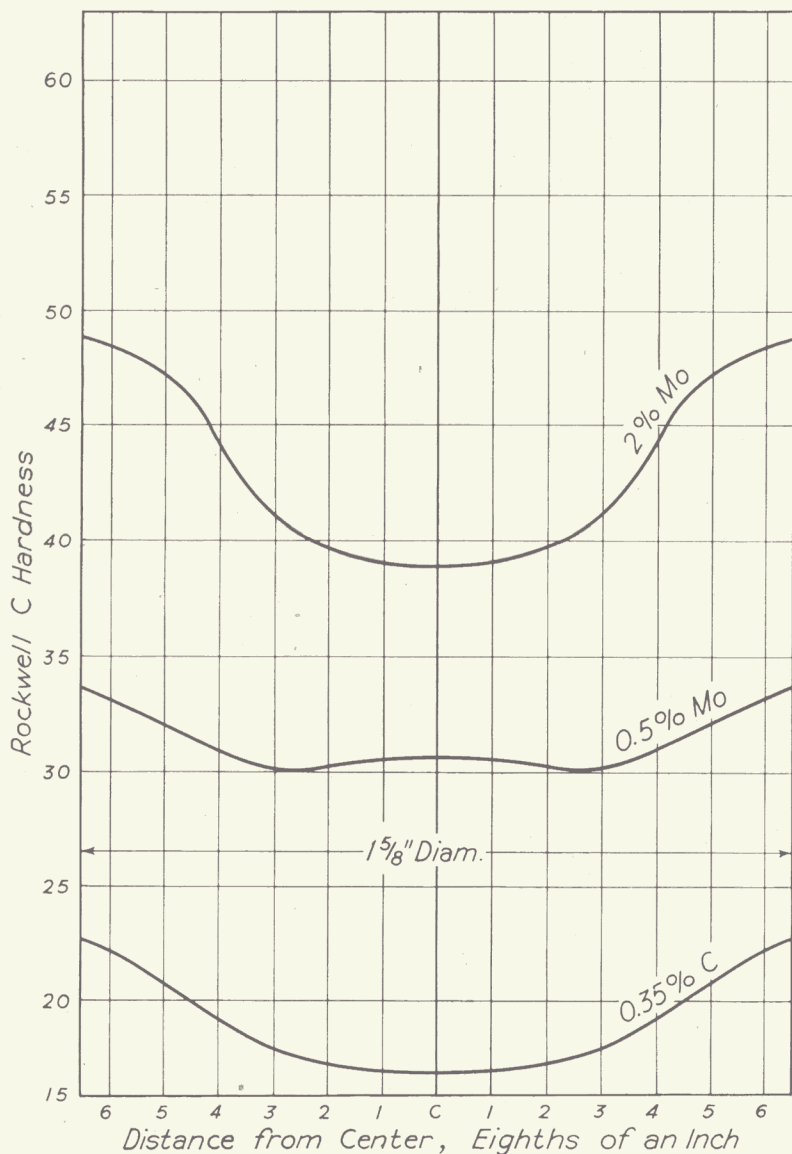


Fig. 106—The Distribution of Hardness across $1\frac{5}{8}$ -Inch Rounds of Three 0.35 Per Cent Carbon Steels with Molybdenum as Indicated. The Bars as Quenched in Oil were Free from Undissolved Carbide. (See Footnote 39).

The further incorporation of about 3 per cent nickel so increases the hardenability that even oil, as well as water, would have resulted in substantially complete martensitization.

Another example of the increase in hardenability due to successively increasing proportions of a dissolved element is

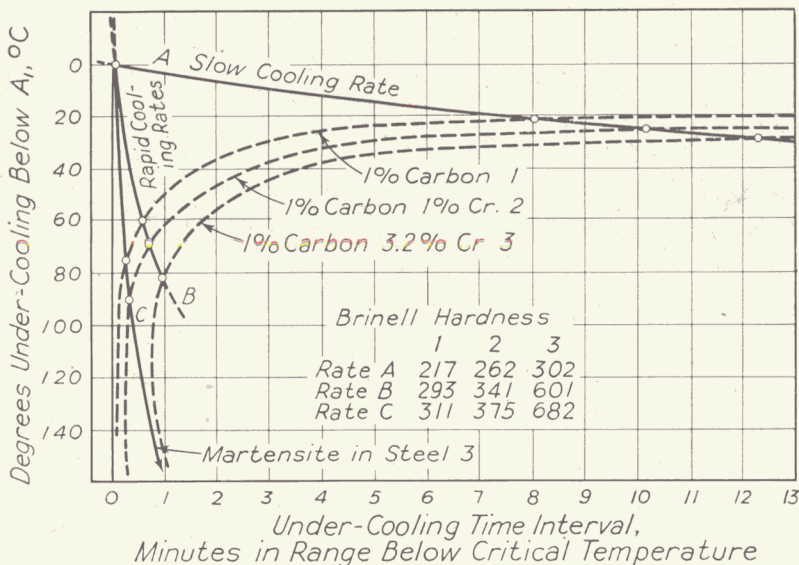


Fig. 107—The Effect of the same Cooling Rates upon the Transformation Temperature of Three Chromium Steels of Like Carbon Content. Degree of Undercooling rather than Actual Temperature is shown in the Ordinate. Time on Linear Scale.

shown in Fig. 105.³⁸ These steels, while representative of the general effect of chromium, were not so quenched as to include an adjustment to precisely constant grain size. The grain size influence would, however, be a minor one.

Coming now to the less easily dissolved elements, it is interesting to note that molybdenum in small proportion exerts a very vigorous effect upon hardenability *when dissolved*. Higher proportions of molybdenum appear not to be

³⁸The data for these hardness distribution curves were secured by J. R. Vilella, L. H. Cooper and H. A. Brown of the United States Steel Corporation Research Laboratory

as effective as would be expected by proportionality; indeed, they are not more effective than chromium. This is illustrated in the 0.35 per cent carbon steels of Fig. 106,³⁹ in which the molybdenum as indicated was substantially all dissolved. These hardness distribution curves indicate that little pure, fully-hard martensite existed in the sections but that the manifestations of hardenability are apparent notwithstanding. Indeed, a fairly reliable estimate of hardenability can be made by certain characteristics of the transformation at the relatively high temperatures which are remote from those producing martensite.

Depression of A_r a Reflection of Hardenability — When any certain type of steel is cooled at some certain rate which depresses the transformation to some 100 to 300 degrees Fahr. below the A_{1c} temperature, its precise undercooling is dependent upon the same fundamentals that determine its hardenability. The undercooling then becomes a measure of hardenability, but some temperature should be selected from which to estimate undercooling. For the same nominal steels this may be taken as constant and therefore ignored entirely only the halt in cooling being observed as to temperature. When more dissimilar steels are compared, the halt point upon slow heating or the temperature inducing the first austenite may be employed as the reference for undercooling.

This is illustrated in any of the three cooling curves of Fig. 107, alike for all three steels, intersected at the transformation temperature, (in terms of undercooling). The curves relate to three different chromium steels of increasing hardenability. The degree of undercooling is seen here to reflect the additions of the deep-hardening element. Note also that the relative hardness itself, though not generally of martensitic range for the steels, also reflects hardenability. One may

³⁹The data for these hardness distribution curves were secured by J. R. Vilella, L. H. Cooper and H. A. Brown of the United States Steel Corporation Research Laboratory.

say that such observations of the depression of the Ar' temperature are *reflections* rather than *measures* of hardenability, as are also hardness determinations of the products of slow cooling, but when employed for a single type of steel by an experienced investigator are useful in predicting hardenability.

Designating Dissolved Alloy versus Hardenability—The most helpful information would be charts or tables for dif-

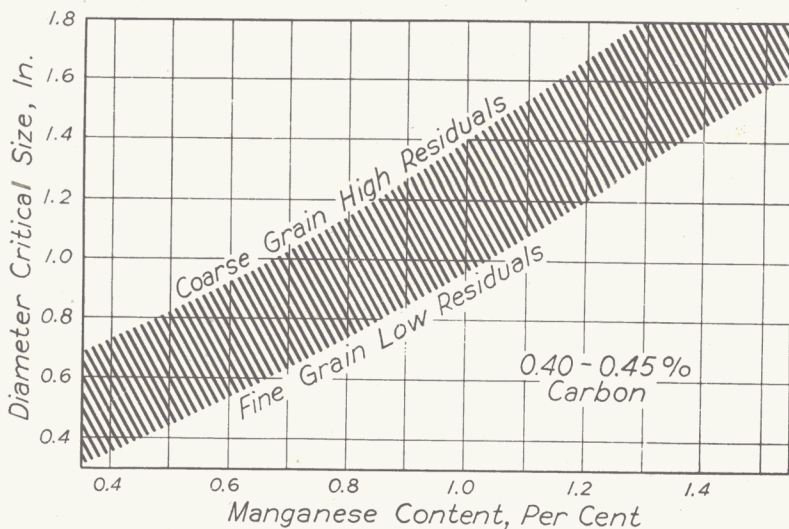


Fig. 108—The Influence of Dissolved Manganese Content upon Critical Size of Round of 0.45 Per Cent Carbon Steels. Water Quench, H = about 5. Slope of Band is probably representative.

ferent brackets of carbon content showing the increase in critical size with increase in alloying element in steels of uniform grain size and homogeneity of the austenite. Such charts are not as yet constructed but Fig. 108 shows in a band the influence of manganese to increase the hardenability (critical size) of 0.45 per cent carbon steels. The quench is a vigorous movement in water with a value of H = about 5. The wide band is employed not so much because the data are inconsistent as because there are not enough based upon

closely corresponding compositions and grain sizes. It is to be expected, however, that data will accumulate so that diagrams of this type will soon be available. If, instead of the critical size for some particular quench, the ideal critical size is plotted, the specific information can then be secured by using the chart of Fig. 98. With respect to manganese the *slope* of the band in Fig. 108 is probably reasonably reliable.

It is an interesting circumstance that the elements dissolved in austenite which most increase its hardenability are those of least solubility, at least in the presence of carbon. Omitting cobalt, nickel is exceedingly soluble and at some 20 to 30 per cent even renders austenite persistent or stable at room temperature, but in comparison with other elements is only a mildly deep-hardening element. Molybdenum and vanadium, with very limited solubility in austenite, are powerful deep hardeners. With some few exceptions the elements are effective almost in inverse relation to the solubilities in austenite, silicon and cobalt, for different reasons standing in somewhat anomalous position.

When present in relatively large proportion, however, the ferrite-soluble, carbide-forming elements contribute relatively less to hardenability if used singly. The most effective additions are combinations of elements from both classes, e.g., the nickel-chromium or nickel-chromium-molybdenum steels.

Elements in Carbides—It has been shown that the proportion of an element which becomes dissolved in austenite is that which is effective in increasing hardenability. The portion of an alloying element combined in undissolved carbide, if not inert, surely plays no helpful part in hardenability; but it has, generally speaking, already exerted an influence in the opposite direction because it has restricted grain growth. It is well to repeat that the heating temperature and interval exert a very strong influence upon this partition of the carbide-forming elements between austenite and the inert, un-

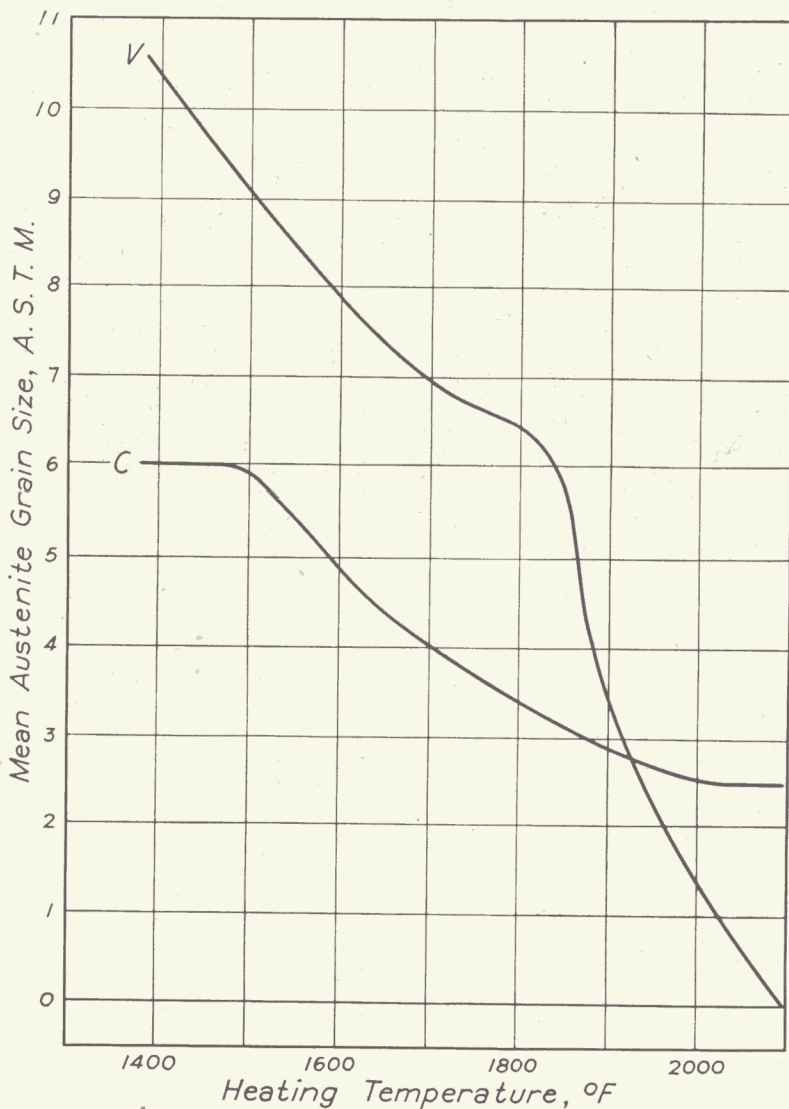


Fig. 109—The Approximate Trends of Mean Grain Size in the Austenite of Two 0.90 Per Cent Carbon Steels alike except for 0.27 Per Cent Vanadium in Steel V. Note Coarsening Ranges. (From Data of Zimmerman, Aborn and Bain).

dissolved carbide. The heating schedule determines which is the dominant role played by these most powerful elements with a dual nature.

The element vanadium is the best one with which to exemplify the carbide effect in hardening, for its influence is more intense than, although similar to, that of tungsten and molybdenum. Titanium is more a carbide-former than vanadium, but with appropriate compositions and temperatures its action is comparable in principle.

The undissolved carbide of vanadium steels robs the austenite of some part of its carbon content and restrains grain growth so that at low heating temperature it generally lowers hardenability. Fig. 109 shows the grain-growth characteristics of two 0.90 per cent carbon steels, C and V, substantially identical except for 0.27 per cent vanadium in steel V. (It will be noted that the extraordinary fineness of the austenite grain requires an extension of the A.S.T.M. Grain Size Numbers beyond No. 8; the same system of numbering is maintained however). It should be noted that there is for both steels a zone of heating temperature in which both steels coarsen at maximum rate with increasing heating temperature.

Consider now the hardness distribution across diameters of the two steels as quenched from successively higher temperatures as shown in Fig. 110.⁴⁰ When the plain carbon steel has achieved a grain size of No. 4 to No. 5 its hardenability is almost precisely equal to that of the vanadium-bearing steel when its grain size is No. 7 to No. 8. It is found that at the temperature inducing this grain size in the vanadium steel, i.e., 1650 degrees Fahr. (900 degrees Cent.), the vanadium-rich carbide is just beginning to be dissolved. Accordingly one may say that the first significant solution

⁴⁰"Some Effects of Small Additions of Vanadium to Eutectoid Steel", by J. G. Zimmerman, R. H. Aborn and E. C. Bain, TRANSACTIONS, American Society for Metals, Vol. 25, 1937, p. 755.

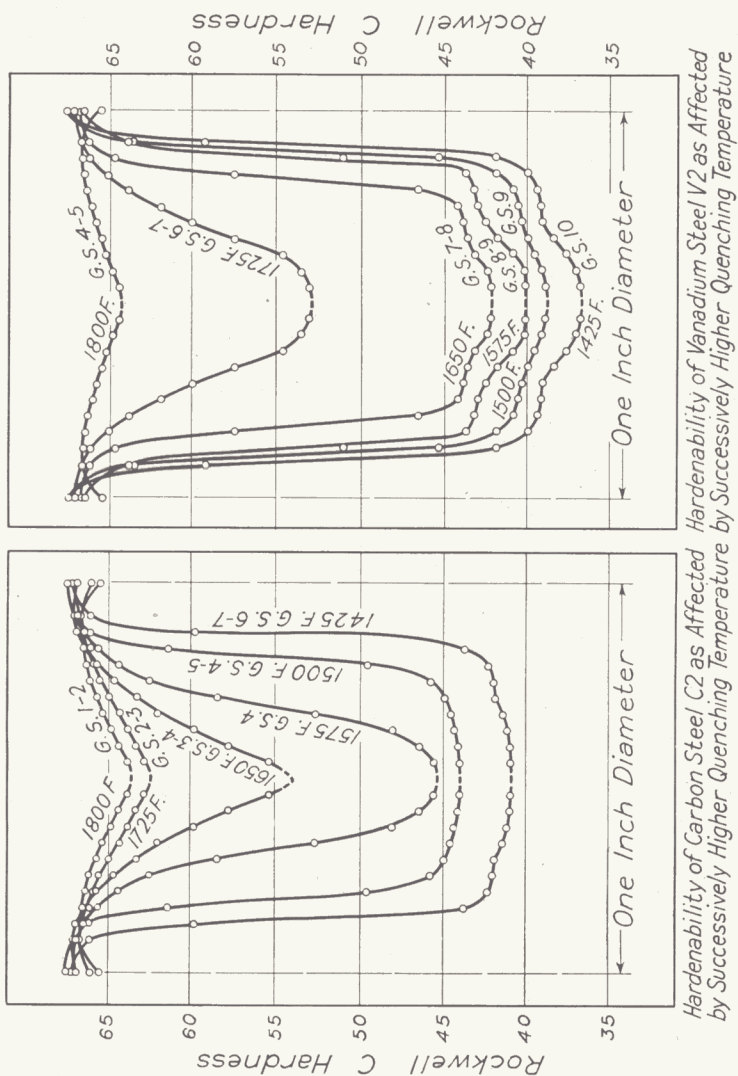


Fig. 110—Hardness Distribution across 1-Inch Rounds of Steels C and V of Fig. 109 as a Function of Grain Size and Heating Temperature. Left, Carbon; Right, Vanadium. (From Data of Zimmerman, Aborn and Bain).

of the vanadium in the austenite has a deep-hardening effect equivalent to a change in grain size from 7-8 to 4-5, or three

A.S.T.M. numbers. Observe further that at the heating temperature at which a further increment of vanadium is dissolved, e.g., 1800 degrees Fahr. (975 degrees Cent.), the steel has acquired a grain size of 4 to 5 and has become exceedingly deep-hardening, and in this respect exceeds that of the plain carbon steel coarsened to a grain size of 1 to 2. Actually a further solution of vanadium (more readily brought about in a somewhat lower carbon steel) induces an extraordinary high hardenability when the grain size approaches or exceeds that of carbon steel; indeed, bars larger than those shown in Fig. 110 must be employed to permit estimation of the increment in hardenability.

It should be stated here that it is possible vanadium may restrict grain growth by virtue of other mechanisms than its contribution of a particular kind of persistent, fine, carbide dispersion. It would be difficult indeed to prove that there were not a dispersion of particles of an oxide or nitride which evades detection. One may say rather that the close parallel between marked grain growth and carbide disappearance in these steels as in tungsten and molybdenum steels (and indeed, carbon steels) makes the present hypothesis satisfactory if not conclusive. Considering the preservation of fine grain after high heating temperature and the accompanying toughness in heat-treated vanadium steels, it is perhaps proper to suggest the use of higher heating temperatures for vanadium steels and similar steels than is often applied. The advantages will be seen also in the subsequent observations on tempering.

To sum up, the carbide-forming elements are natural grain-growth restrainers so long as some proportion of the fine carbide particles remains. As dissolved, they are very powerful deep-hardening elements. Advantage may be taken of both good characteristics by suitable choice of heating schedule.

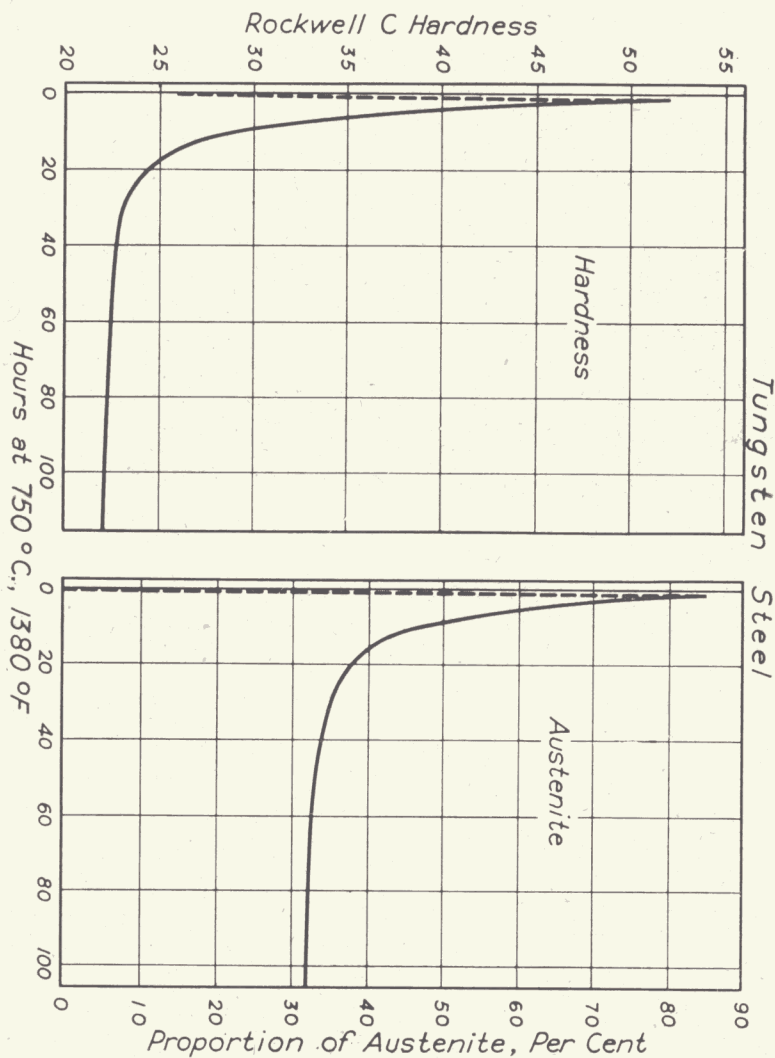


Fig. 111—Left—Change in Hardness Developed upon Quenching a 0.55 Per Cent Carbon, 2 Per Cent Tungsten Steel as a Function of Heating Time at Relatively Low Temperature. Right—Change in Proportion of Austenite. (See Footnote 41).

Anomalous Hardening at Low Heating Temperatures—

To employ a valuable carbide-forming element only as a source of hard inert particles is justifiable for the sake of

abrasion resistance as needed in cutting tools, provided sufficient carbon and alloy are present to form an austenite of high hardenability and furnish needed strength in the resulting tempered martensite. When, however, heating temperatures are employed which are so low as to develop only a low-carbon, low-alloy austenite composition, the final properties are rarely such as would be demanded for any purpose. In so handling the steels with tungsten, molybdenum, vanadium, and to some extent chromium, poor use is made of the alloy content and some rather surprising behaviors may be encountered.

Consider a 2 per cent tungsten steel with about 0.55 per cent carbon. At equilibrium, such a composition is actually composed of ferrite, tungsten-rich carbide and a little austenite at a temperature of 1380 degrees Fahr. (750 degrees Cent.). But this was not always known. In the course of developing some information on equilibrium conditions such a steel was heated for a number of increasing intervals at this temperature.⁴¹ A short heating followed by quenching produced a hardness of 52 Rockwell C, but increasing intervals of heating did not increase the resulting hardness, instead longer heating gradually reduced the as-quenched hardness. The behavior is shown in Fig. 111.

Having regarded the increase of heating period as a means of dissolving more carbon and alloy this was a perplexing circumstance. That it was not a matter of decarburization was verified and in the course of microscopic observation it was found that the proportion of austenite formed at the heating temperature was actually falling off with lapse of time. The unexpected behavior demanded an explanation and it became quite apparent that the unique circumstance here was the low heating temperature.

Presumably at this low heating temperature, carbon may

⁴¹Unpublished work of Rutherford, Aborn and Bain.

migrate with moderate velocity in terms of microscopic distances. Accordingly, a considerable proportion of austenite of rather low carbon content is first formed. Later, much later, the tungsten migrates and asserts its influence to form ferrite with only a little austenite of higher carbon content and a certain amount of tungsten-rich carbide.

The carbide-forming elements, when actually dissolved, exert a far greater effect upon hardenability than do corresponding proportions of the elements manganese, nickel and silicon. Furthermore, this retardation of the reaction rate for austenite transformation is mainly restricted to the upper temperatures of the S-Curve; accordingly in spite of their deep-hardening nature additions of the carbide-forming elements do not cause the preservation of much austenite after the quench. Chromium, which is a milder carbide-forming element, does foster some austenite retention but far less than corresponding amounts of manganese.

The interest at this point is in the undissolved carbide-forming elements and while the effect upon grain growth is marked, and the hard residual carbides have special value in tools, the most interesting of all their functions is revealed only in tempering. Accordingly these elements will be brought up for discussion again.

Elements in Nonmetallic Inclusions — The influence of nonmetallic inclusions is presumably much the same in alloy steels as it is in compositions more nearly approaching the simple iron-carbon alloys and this effect has already been discussed. It is, however, inevitably obscured in the presence of carbide-forming elements. Particular attention was given to the very effective additions of aluminum which, presumably, operate through alumina, aluminum nitride or other aluminum-bearing inclusions. The aluminum treatment, when skillfully utilized, is of course applicable to the elevation of grain-coarsening temperature in practically all steels

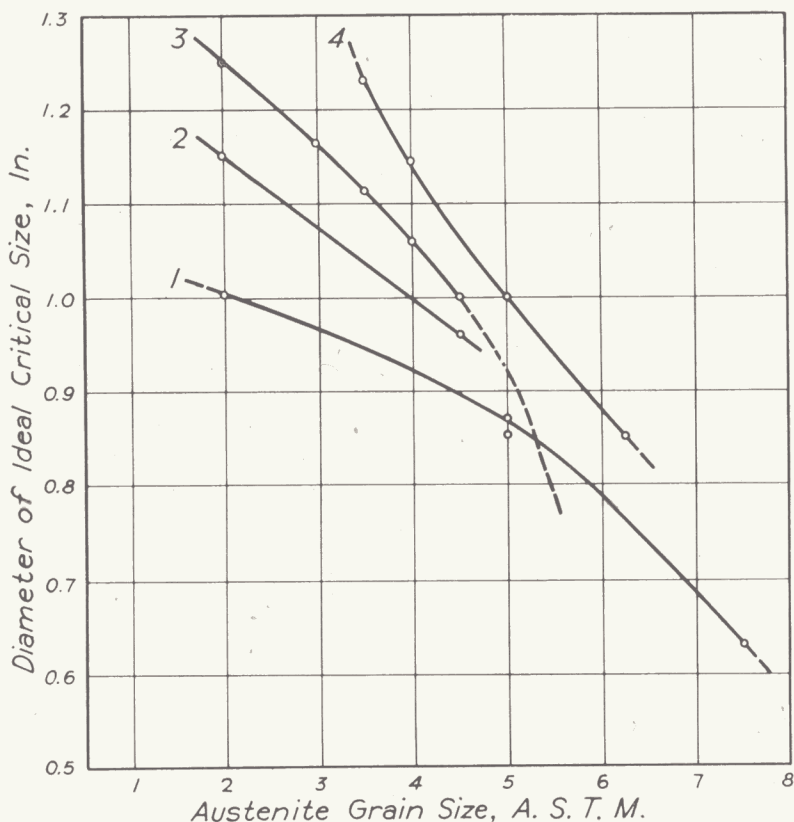


Fig. 112--Ideal Critical Size in Four Compositions as Changed by Austenitic Grain Size at Quenching Temperature.

other than high-aluminum or rimming compositions and has been so used in many compositions.

It is worthy of repetition that the dispersion, whatever it may be, of an aluminum compound may act not only to restrict austenitic grain growth, and thereby lessen hardenability at normal heating temperatures, but may act directly by stimulating nucleation. This is evident in typical "fine-grained" steels (with high coarsening temperature) which, however, have had their austenite grain greatly coarsened by excessive heating. Under certain circumstances such coars-

ened steels have the low hardenability of fine-grained steels with their tough centers. This circumstance has been described by the author and need not be considered in detail here. It may, however, not be out of place here to consider the magnitude of the change in hardenability brought about by coarsening of the austenite grain size under circumstances

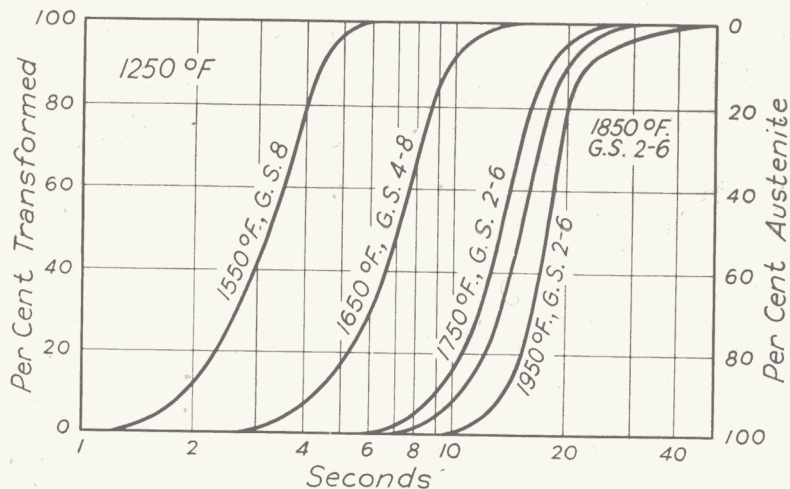


Fig. 113—Progress of Reaction,—Austenite to Pearlite, as Retarded by Coarser Mean Austenite Grain Size. Reaction at 1250 Degrees Fahr. (675 Degrees Cent.).

in which it appears no other influences are significant, — such as homogeneity of austenite, undissolved carbide, particle nucleation, etc. Fig. 112 shows the probable effect of austenitic grain size upon the absolute hardenability in terms of ideal critical size in four steels of the following approximate analyses:

1. 0.40 per cent carbon, 0.80 per cent manganese
2. 0.42 per cent carbon, 0.90 per cent manganese
3. 0.75 per cent carbon, 0.30 per cent manganese
4. 0.90 per cent carbon, 0.30 per cent manganese

The falling off of hardenability of steel 3 at finer grain size is thought to be abnormally great because of undissolved car-

bide resulting from inadequate heating. While these examples are not as uniform as might have been hoped for they indicate in a most general way that in such steels a decrease in grain size number of three may perhaps alter the ideal critical size by about a quarter of an inch. The general influence might have been illustrated in nickel, chromium, or higher manganese steels, had suitable data been available, but in steels wherein carbide solution and subsequent homogenization is difficult the single factor of grain size would not have been brought out free from other influences. As a matter of possible interest the fundamental transformation rate at higher temperature for different mean sizes of austenite grain is shown in Fig. 113.

The nonmetallic inclusions which are beneficial in steel are those which aid in retaining a fine austenite grain size or nucleate martensite formation, and thereby contribute toughness. It is of no great benefit if they merely reduce hardenability alone. The characteristics of such nonmetallics are these:

1. Formation of easily eliminated inclusions in the liquid steel, probably by acquiring large size.
2. Relatively high solubility just below the solidification temperature and low solubility at heat-treating temperature.
3. Low diffusivity, at the temperature of falling solubility, of at least one of the elements entering into the compound of the inclusion.
4. A capacity for stimulating austenite nuclei upon heating or possibly martensite nuclei upon quenching.

These characteristics probably insure the absence of injuriously large inclusions and provide the properties which are regarded as valuable and to which the name "body" was at one time applied. Some agents are known which appear to act in these ways (in addition to the carbide particles which are in part similar), though there is question whether they are quite as effective as aluminum.

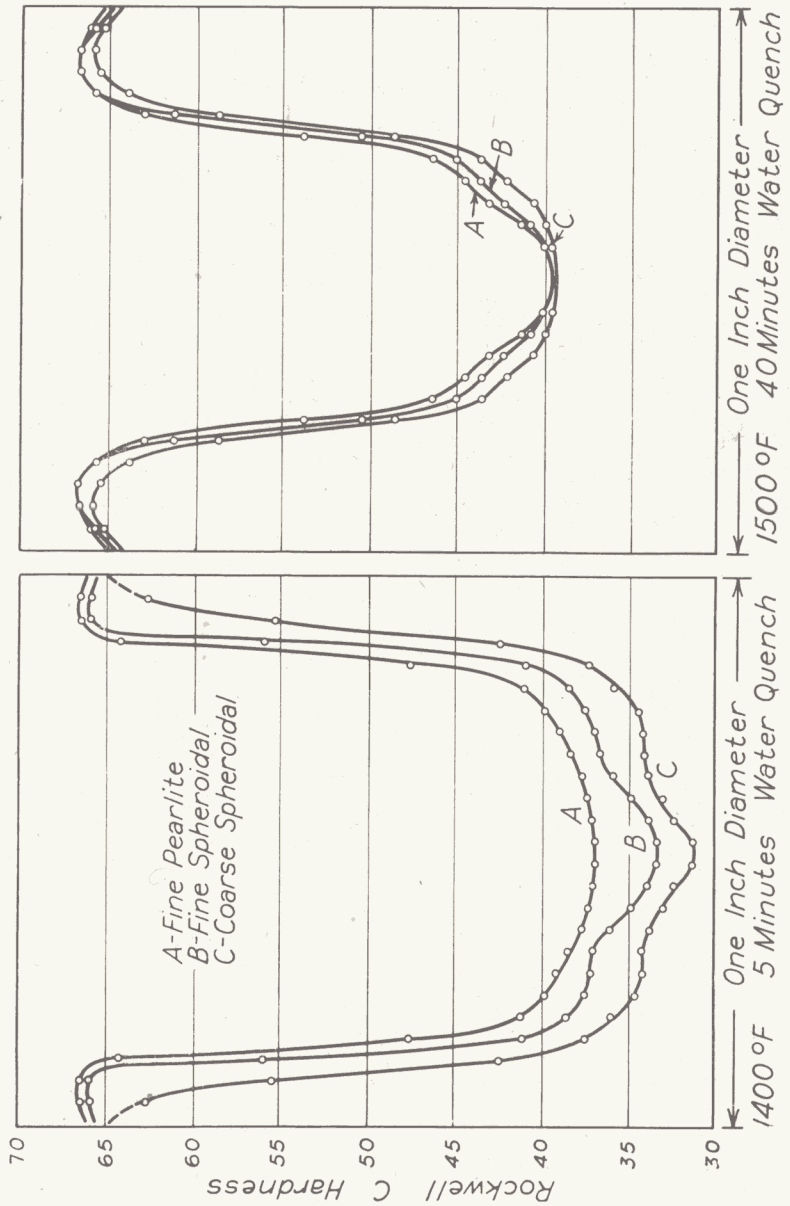


Fig. 114—Hardness Distribution in 1-Inch Rounds of Eutectoid Steel as Influenced by Structure Entering Heating Bath for a Constant Heating Schedule. Left—Short Time at 1400 Degrees Fahr.; Right—Longer Time at 1500 Degrees Fahr. (See Footnote 42).

Microscopic Homogeneity and Hardenability — It has been shown that the solution of carbide is a time-consuming operation; even though it is rapid in pure iron-carbon alloys it is slow in some alloy steels. This is because diffusion itself is time-consuming and only through diffusion may austenite be rendered homogeneous, for the austenite formed at the former location of carbide particles is very rich in carbon. The author has seen quenched steels which had been heated for hardening from a coarsely pearlitic state in which the markings of the original lamellae were clearly visible in the microscopic specimen of resulting martensite. The reason the etching agent may reveal the pattern is that the martensite was of uneven carbon content; low in the former ferrite lamellae zones, high in the cementite lamellae locations.

An inhomogeneous austenite is not unlike an intimate mixture of two steels (and intermediate ones also). The high carbon and alloy regions of microscopic extent have high hardenability, the remainder low. For pure martensite the steel must be quenched at the required critical rate of the leanest austenite. For practical purposes hardenability is impaired by inhomogeneity resulting from insufficient heating.

Clearly the simplest way in which to render homogenization easy is to form austenite from the finest possible dispersion of carbide compatible with other requirements such as machinability. Thus the structure prior to heat-treatment, especially when minimum heating time and temperature is employed, has a marked influence upon hardenability. Double heat-treatment, first, from high temperature for homogenization and, second, at lower temperature for fine grain size and toughness, is exceedingly desirable and a commendable practice.

The influence of the structure entering the heating bath (for a constant schedule) upon hardenability is well illus-

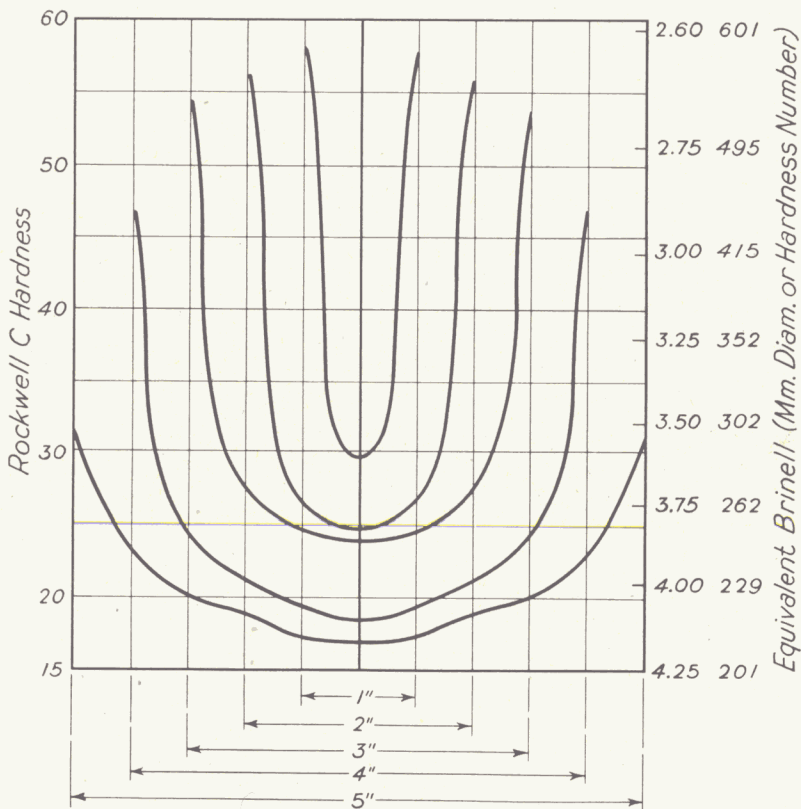


Fig. 115—Hardness Distribution in Quenched Round Bars of S.A.E. 1040 Steel. Vigorous Water Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

trated in Fig. 114.⁴² It will be noted that a short heating period (left) fails to equalize the three conditions of the steel, whereas a longer interval (right) at high temperature nearly does so. The three structures were a normally coarsely spheroidized state, a more finely spheroidized one and a fine pearlite produced by a mild oil quench. The heating was such as to just barely dissolve the carbide and produce essentially the same grain size in all.

⁴²Unpublished work of Porter R. Wray, United States Steel Corporation Research Laboratory.

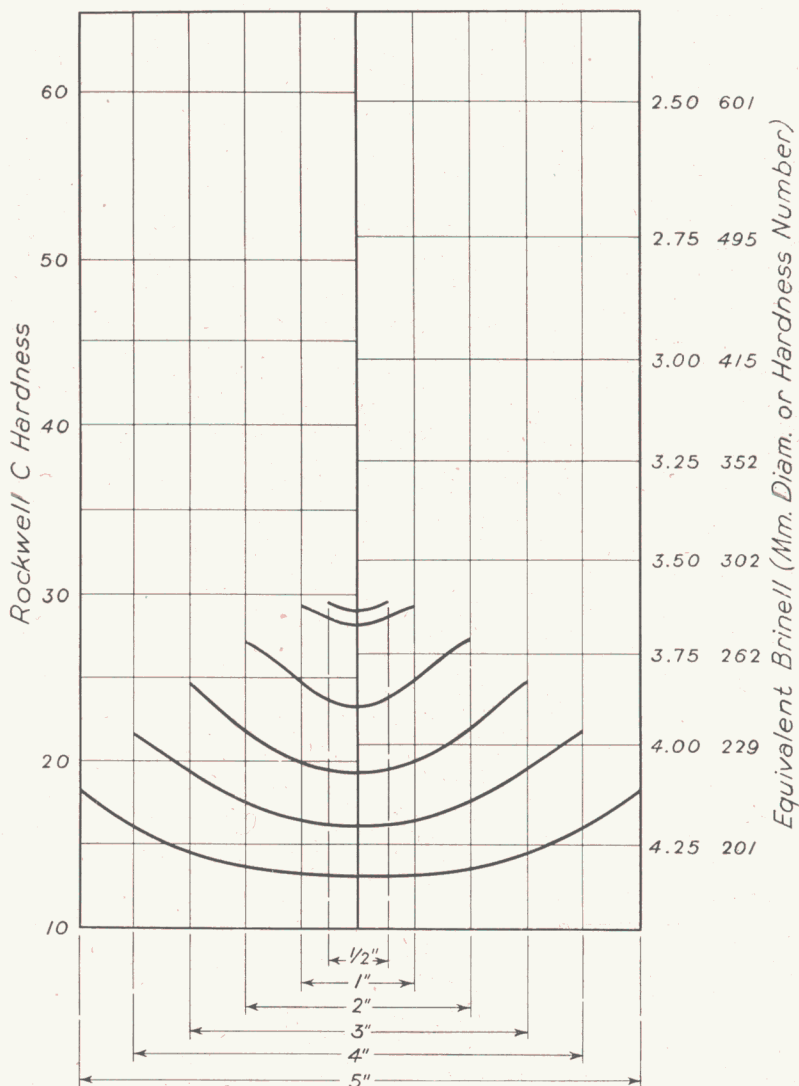


Fig. 116—Hardness Distribution in Quenched Round Bars of S.A.E. 1040 Steel. Average Oil Quench from 1525 Degrees Fahr. (Carilloy Steels).

Hardenability of Familiar Compositions — In order to gain some familiarity with the hardenability contributed by

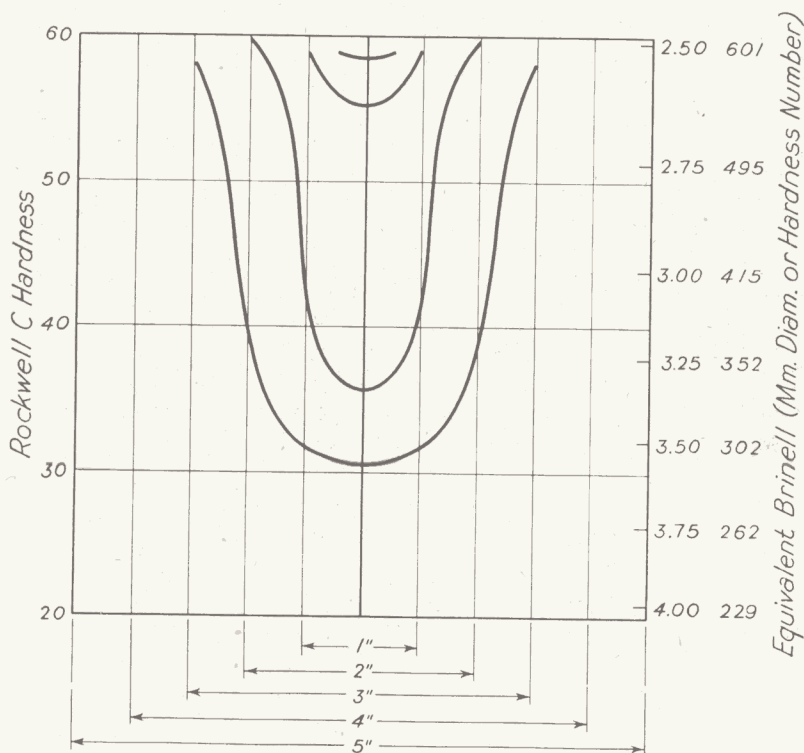


Fig. 117—Hardness Distribution in Quenched Round Bars of 0.40-0.45 Per Cent Carbon Amola Steel. Vigorous Water Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

the alloying elements in well known steels it may be well to examine the hardness distribution across mean diameters of rounds of various sizes. It will perhaps be noted that for this purpose of gaining acquaintance with various steels in respect to hardenability one has an inclination to turn to the scheme of transverse hardness distribution. If so, this argues for its suitability for the steels of low and intermediate hardenability.

In Figs. 115 to 132 a number of familiar steels are portrayed in a manner to indicate at a glance the effect of size of round upon the hardness secured at various depths after

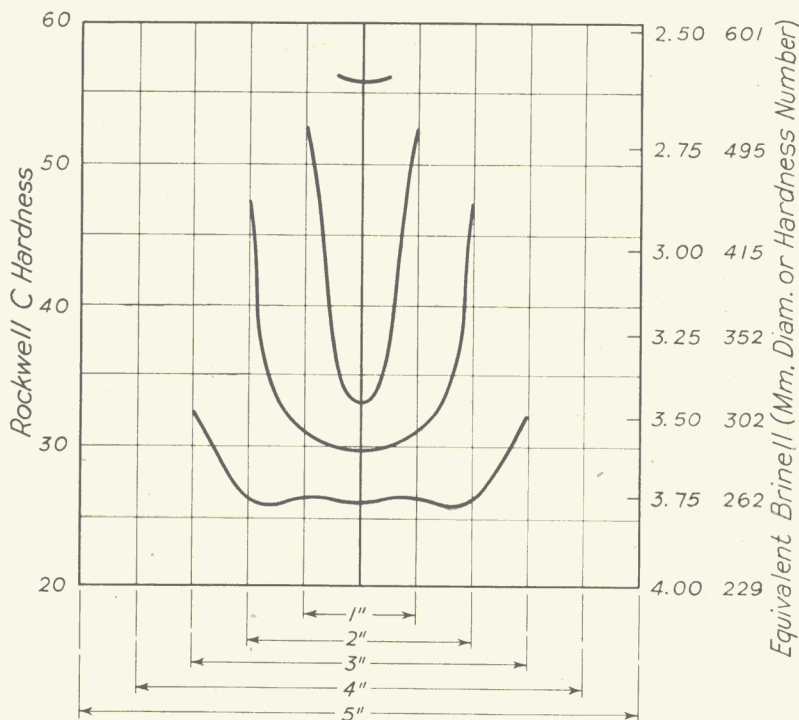


Fig. 118—Hardness Distribution in Quenched Round Bars of 0.40-0.45 Per Cent Carbon Amola Steel. Vigorous Oil Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

the indicated quench, oil or water. While these are probably fairly representative of what may, in general, be secured, it will be understood that some variations will exist among heats of steel and particularly among different heating and quenching techniques. The heating temperature is also specified although this figure alone will not identify a unique degree of solution or homogenization of the alloying elements. In the case of the vanadium, and possibly also the molybdenum-bearing steels, some little carbide probably remained undissolved. A very considerable disparity may exist between the results secured at different hardening shops in the instance

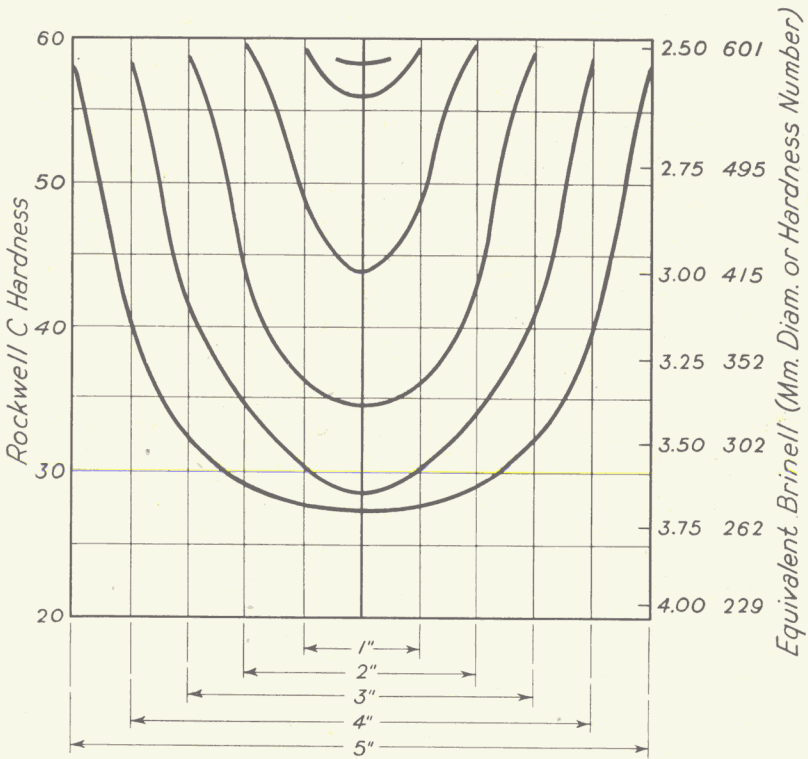


Fig. 119—Hardness Distribution in Quenched Round Bars of S.A.E. T1340 Steel. Average Water Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

of steels carrying these elements because of their great contribution to hardenability in small proportion when dissolved, and because the amount dissolved depends so closely upon the heating schedule. The structure entering the heating furnace has also an influence upon alloy solution.

The steels shown in the hardenability charts of Figs. 115 to 132, inclusive, exhibit a continuous sequence of increasing depth of hardening. The next table shows also, for convenience, the approximate ideal critical size (diameter of round just half martensitic at center after the ideal quench)

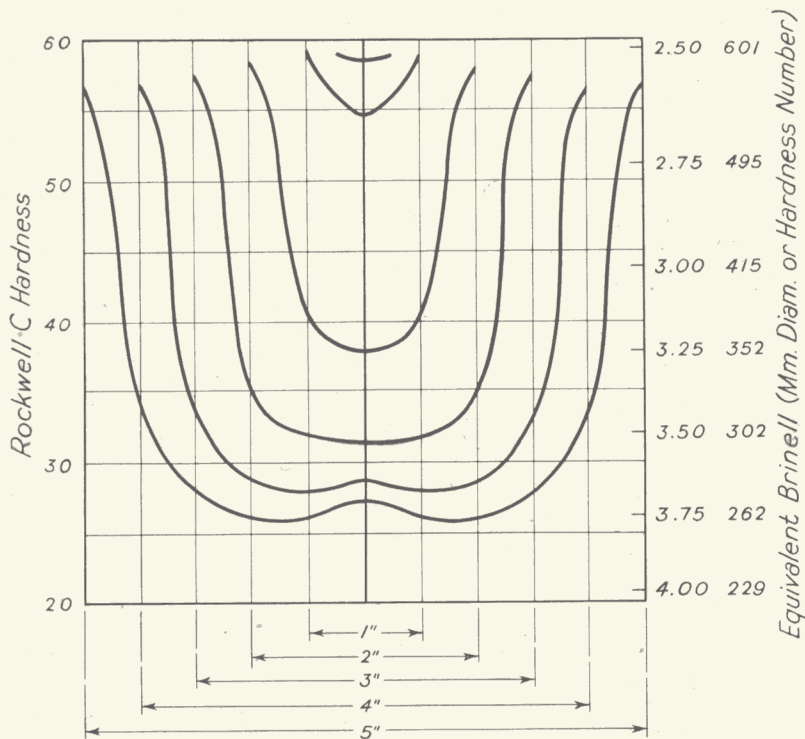


Fig. 120—Hardness Distribution in Quenched Round Bars of S.A.E. 5140 Steel. Average Water Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

for a number of the particular steels as handled in the present observations, which, however, corresponds generally to the schedules of heating often applied.

Steel Designation	Carbon	Manganese	Silicon	Nickel	Chromium	Molybdenum	Ideal Critical Size. In.
1040-45	0.40-0.45	0.60-0.90	0.15-0.30	0.9-1.3
Amola	0.40-0.45	0.70-0.90	0.20-0.30	0.15-0.25	1.3-1.9
T1340-45	0.40-0.45	1.60-1.90	0.15-0.30	1.7-2.3
5140	0.40-0.45	0.60-0.90	0.15-0.30	0.80-1.10	2.2-2.9
5140-45	0.40-0.45	0.60-0.90	0.15-0.30	1.0-1.50	0.45-0.75	2.3-3.0
6140-45*	0.40-0.45	0.60-0.90	0.15-0.30	0.80-1.10	2.4-3.1
2340-45	0.40-0.45	0.60-0.90	0.15-0.30	3.25-3.75	2.8-3.6
4140	0.40-0.45	0.60-0.90	0.15-0.30	0.80-1.10	0.15-0.25	3.0-4.0
3240-45	0.40-0.45	0.30-0.60	0.15-0.30	1.5-2.0	0.90-1.25	3.4-4.2
3340	0.40-0.45	0.30-0.60	0.15-0.30	3.25-3.75	1.25-1.75	7.0-9.0?

*Vanadium content 0.15 minimum.

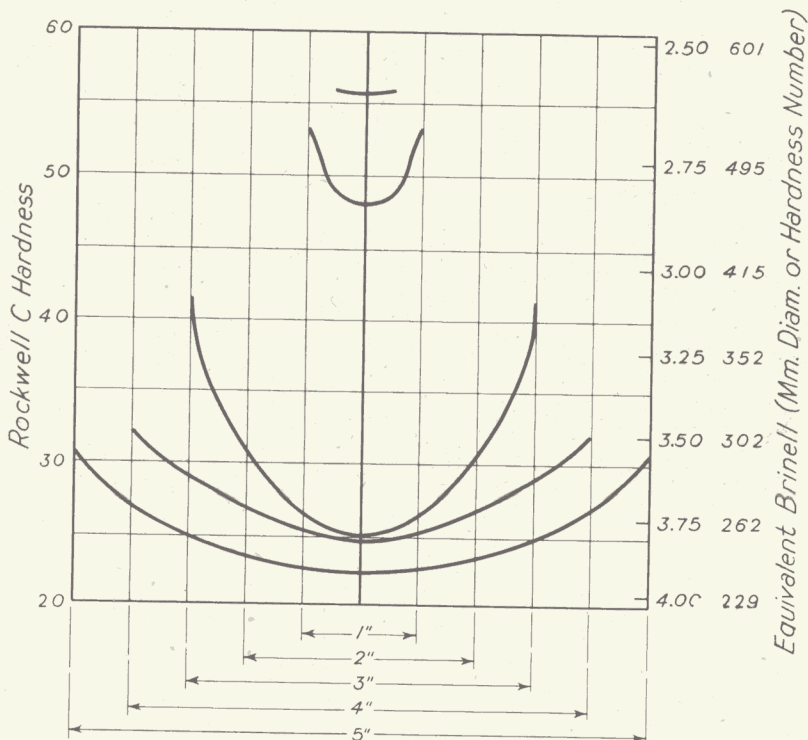


Fig. 121—Hardness Distribution in Quenched Round Bars of S.A.E. 5140 Steel. Average Oil Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

Extra-Equilibrium Constitution Diagrams — In the course of study of some of the iron-carbon-alloy systems some investigators have recorded the structures secured after the same cooling rate from the same high temperature for all the several compositions investigated. This information is then conveniently reported in the form of three component diagrams basically like those of Figs. 47 to 50, which will then be marked off into regions of like microstructure, e.g., martensite, pearlite, etc. To insure that a constant cooling rate actually was impressed upon all specimens, the structure usually is examined at a constant location within each

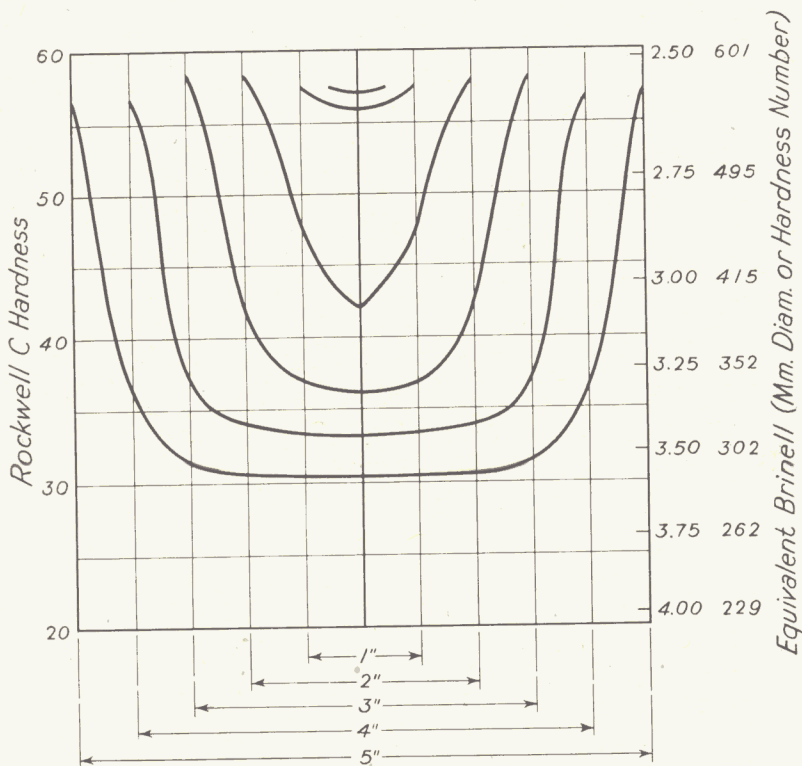


Fig. 122—Hardness Distribution in Quenched Round Bars of S.A.E. 3140 Steel. Average Water Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

similar piece. The cooling may be that of a quench or a mere holding in air. Such diagrams are then informative as to the contribution to hardenability made by the alloying element, that is, the portion of alloy content which is dissolved.

Diagrams of this type are available for the manganese, chromium, molybdenum and nickel alloys and are set forth in Figs. 133, 134, 135, 136 and 137. It is to be hoped that further systems will be reported in this way, making available diagrams which may replace the inaccurate ones of this type, attributed to Guillet, which are scarcely even schematically correct.

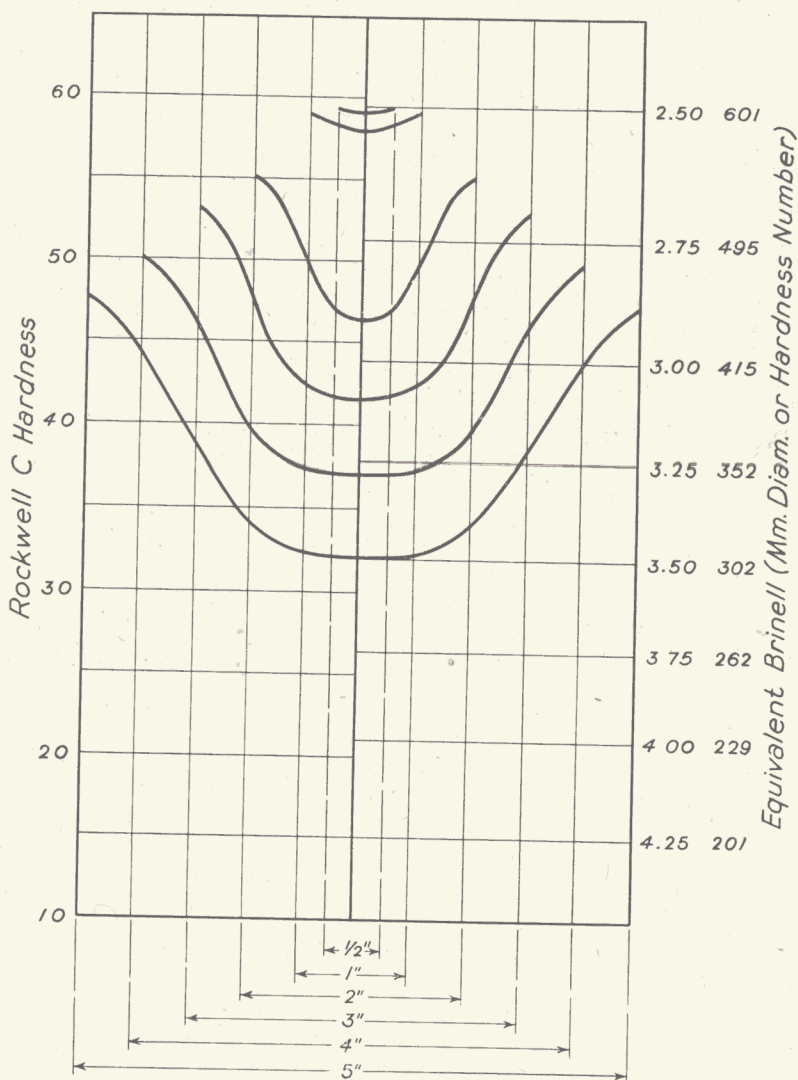


Fig. 123—Hardness Distribution in Quenched Round Bars of S.A.E. 6140 Steel, Average Water Quench. (Carilloy Steels).

Factors Extraneous to the Steel — Thus far consideration has been given very strictly to the hardenability of a steel as it is just ready to be quenched. It will be understood

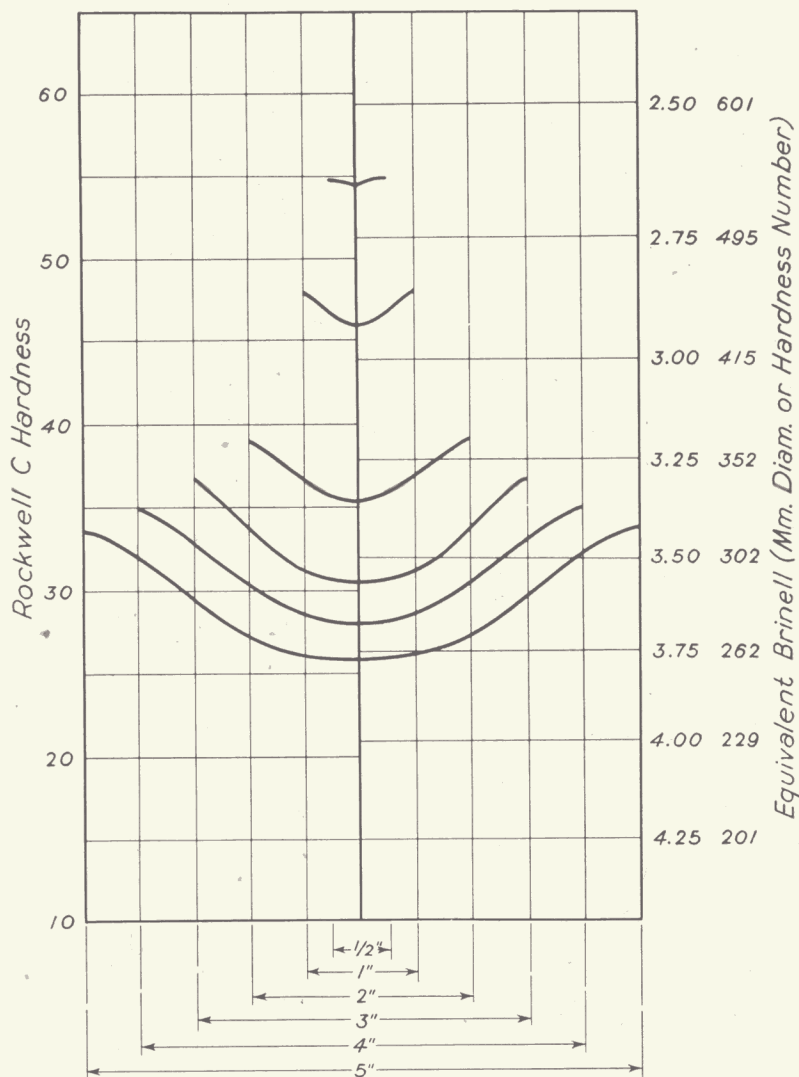


Fig. 124—Hardness Distribution in Quenched Round Bars of S.A.E. 6140 Steel. Average Oil Quench. (Carilloy Steels).

that the mode of handling the steel thereafter does not change its hardenability; indeed, it is the purpose of the analysis of Grossmann *et al.* to separate completely this property of the

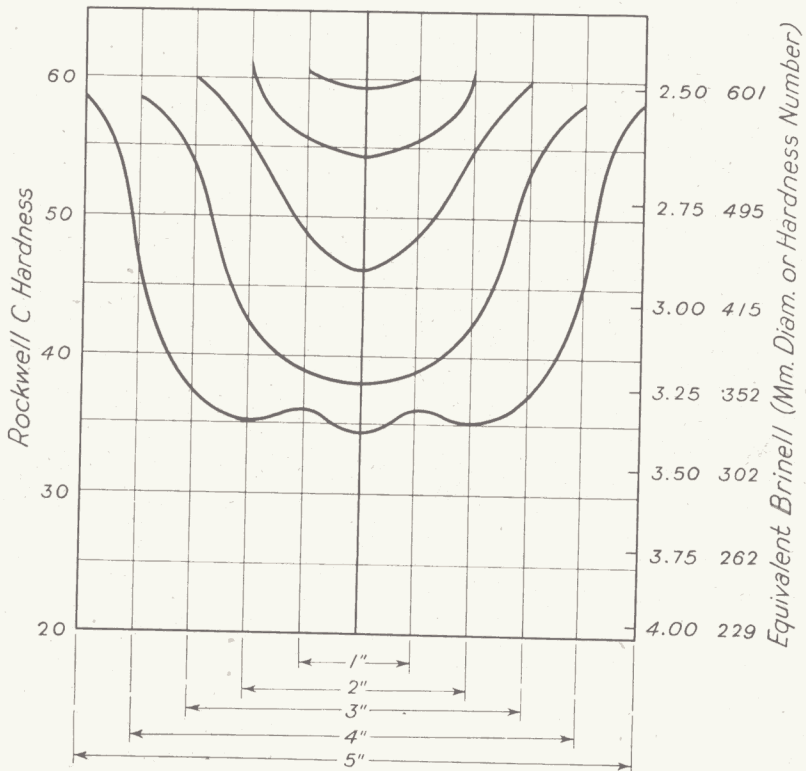


Fig. 125—Hardness Distribution in Quenched Round Bars of S.A.E. 2340 Steel. Moderate Water Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

steel from the conditions of the quench, so that the former may be expressed as a single measure, i.e., the ideal critical size, and the latter by the factor H . So long as the conditions of quenching are unchanged from one specimen to another when two suitably sized bars of homogeneous steel are quenched, both the effective quenching velocity and the hardenability of the steel may be evaluated. This is, however, not as easy to realize in practice as might be supposed. It will be well to consider here one way in which discrepancies may enter an investigation of hardenability.

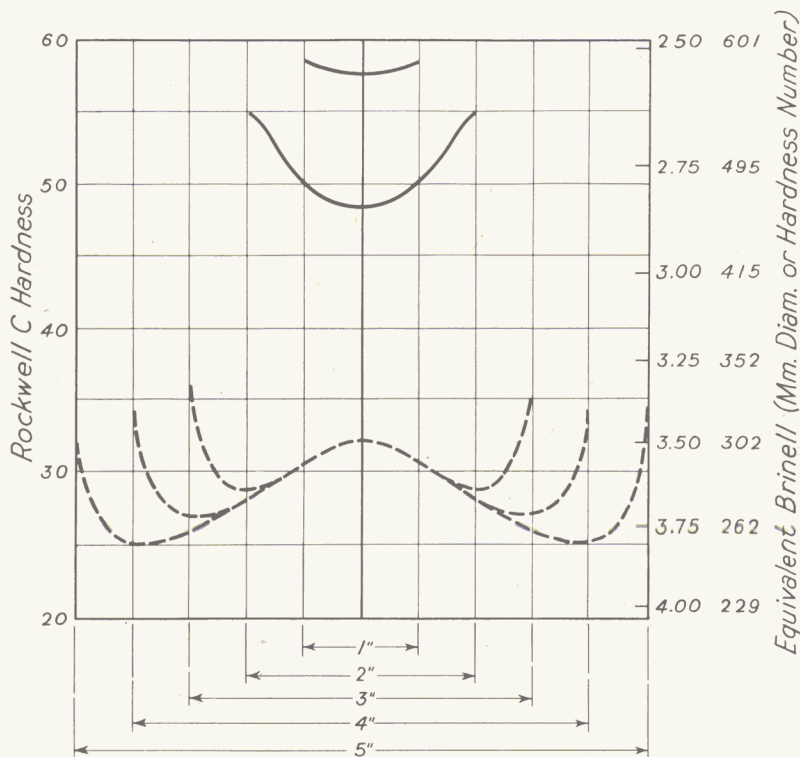


Fig. 126—Hardness Distribution in Quenched Round Bars of S.A.E. 2340 Steel. Average Oil Quench from 1525 Degrees Fahr. (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corporation).

Suppose a certain quenching bath has been used in connection with a certain heating furnace and its effective heat abstracting characteristics have been well established for the work usually handled. Let us assume that it shows that $H = 4.0$. Now the same steels may be heated in other furnaces or in baths and again quenched into the same bath. Instead of a mildly protective atmosphere in a furnace, air may be present in an electric muffle furnace with a considerable convection current. Or the specimens may be heated in salt or lead, reducing the possibility for oxidation.

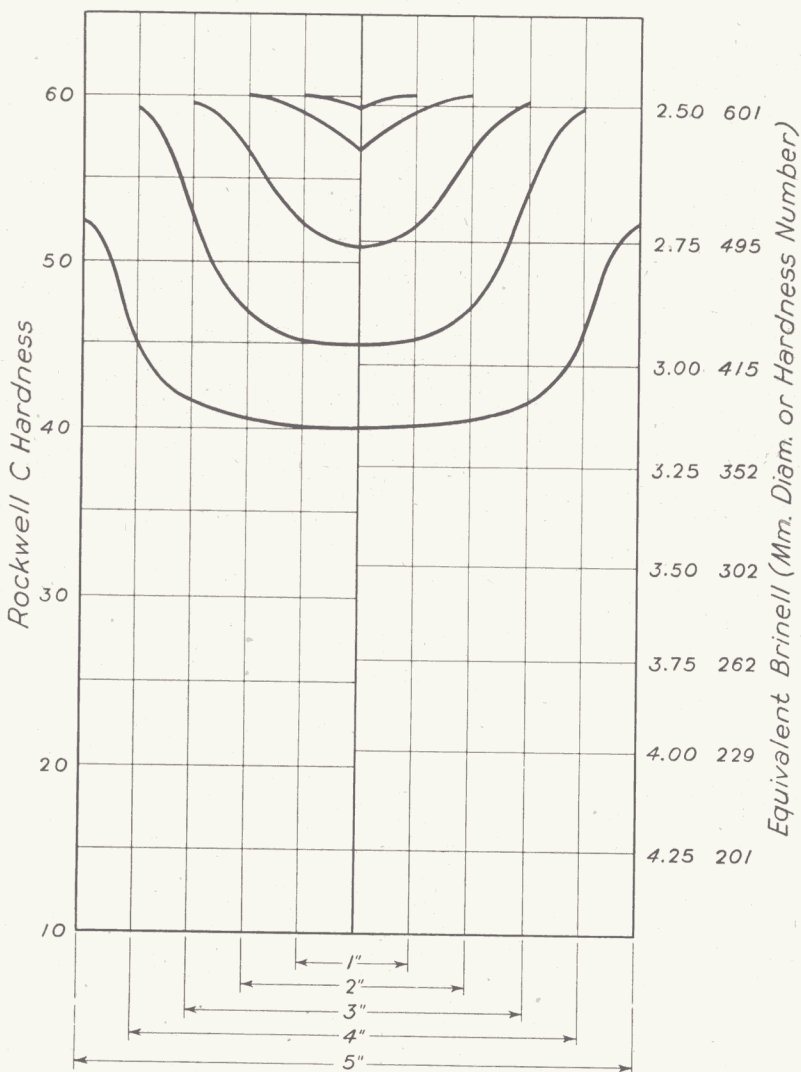


Fig. 127—Hardness Distribution in Quenched Round Bars of S.A.E. 4140 Steel. Average Water Quench. (Carilloy Steels).

Fig. 138 shows the distribution of hardness⁴³ across the 1-inch specimens from the same bar of steel (0.75 per cent

⁴³The data employed in Figs. 138, 139 and 140 were secured by G. E. Guellich of the United States Steel Corporation Research Laboratory.

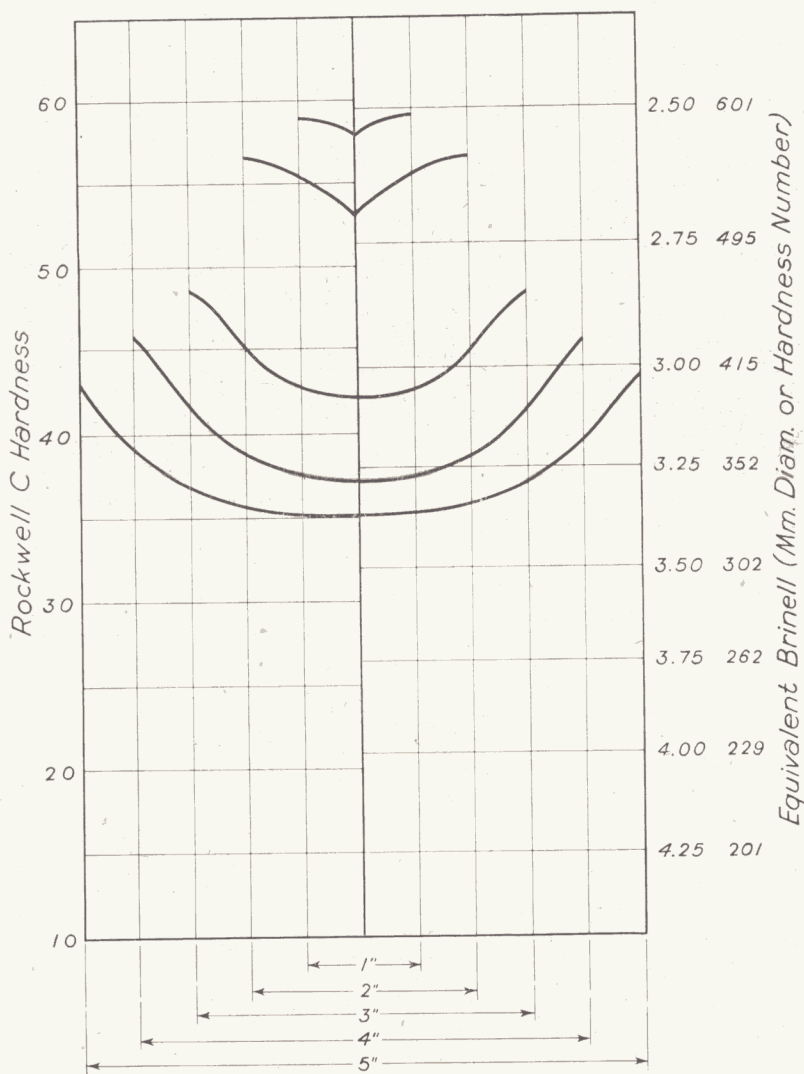


Fig. 128—Hardness Distribution in Quenched Round Bars of S.A.E. 4140 Steel. Average Oil Quench. (Carilloy Steels).

carbon, 0.40 per cent manganese) heated to induce equal grain size and equal homogenization. They are quenched into the same bath with the same agitation. The difference,

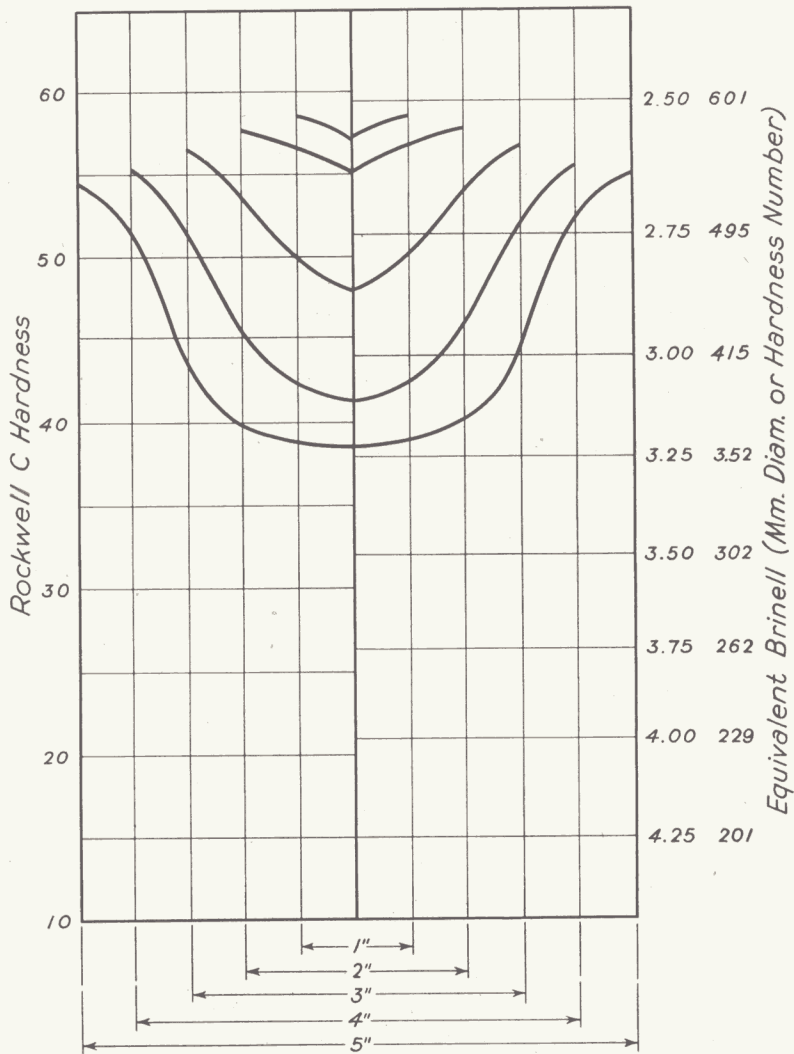


Fig. 129—Hardness Distribution in Quenched Round Bars of S.A.E. 3240 Steel. Average Water Quench. (*Carilloy Steels*).

however, which induced the several hardness distribution curves, was that of the surface itself; i.e., the amount and nature of scale. Note the high *apparent* hardenability of the

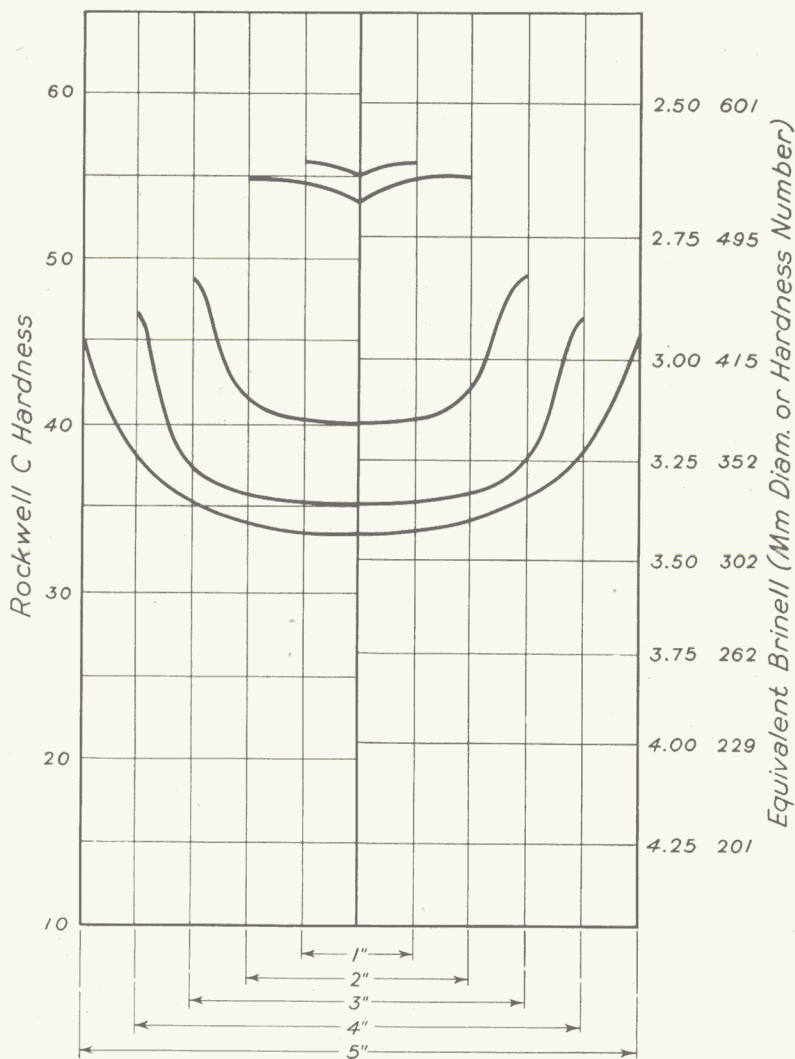


Fig. 130—Hardness Distribution in Quenched Round Bars of S.A.E. 3240 Steel. Average Oil Quench. (*Carilloy Steels*).

specimen quenched with a bright surface from a well deoxidized lead bath. Now, the steel itself did not suffer a change in hardenability, — but the individual specimens did, because

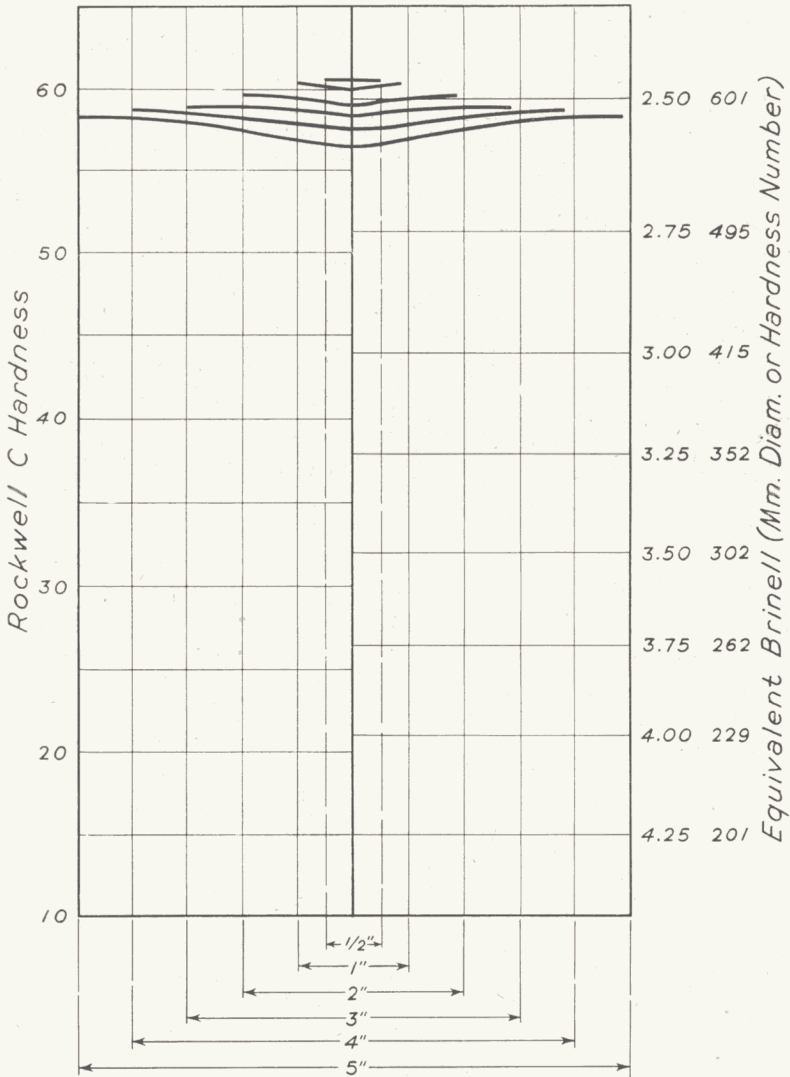


Fig. 131—Hardness Distribution in Quenched Round Bars of S.A.E. 3340 Steel. Average Water Quench. (Carilloy Steels).

the same quenching technique failed to cause equal removal of heat in the four instances.

It would seem most logical to regard that the steel

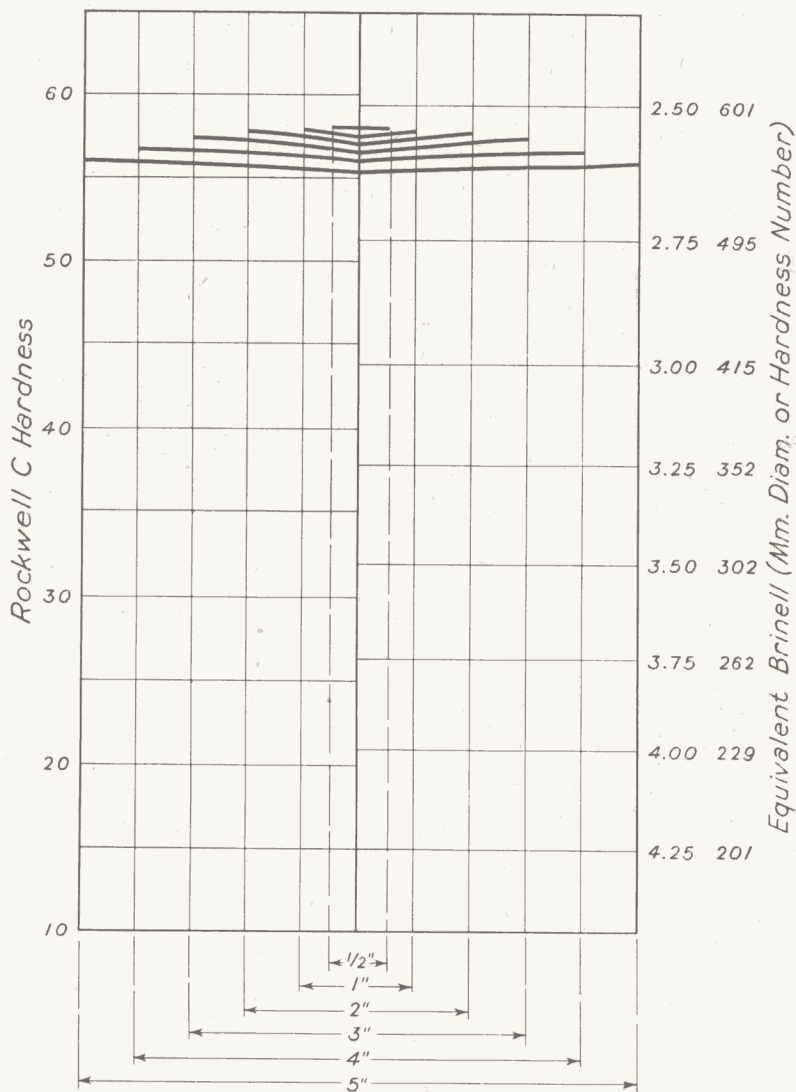


Fig. 132—Hardness Distribution in Quenched Round Bars of S.A.E. 3340 Steel. Average Oil Quench. (Carilloy Steels).

remained constant and that the H value changed. If one makes this choice in allocating the surface condition it merely means that, in studies of hardenability, the heating technique

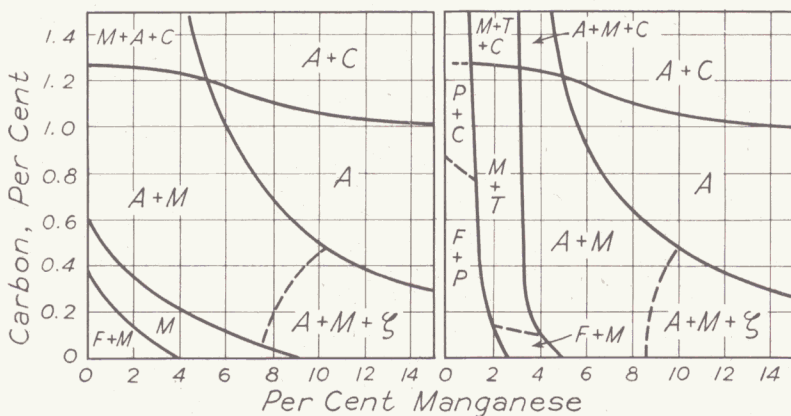


Fig. 133—The Extra-Equilibrium Constituents Developed in the Iron, Manganese, Carbon System as (Left) Quenched and (Right) Air-Cooled in $\frac{1}{2}$ -Inch Cubes. (Bain, Davenport and Waring).

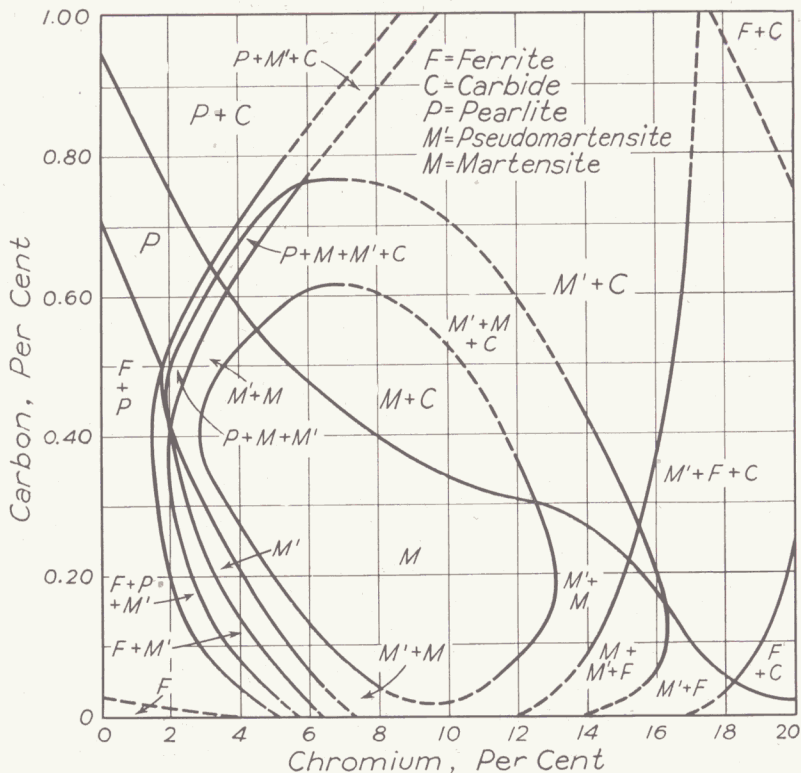


Fig. 134—The Extra-Equilibrium Constituents Developed in the Normalized Iron, Chromium, Carbon System. (Union Carbide and Carbon Laboratories).

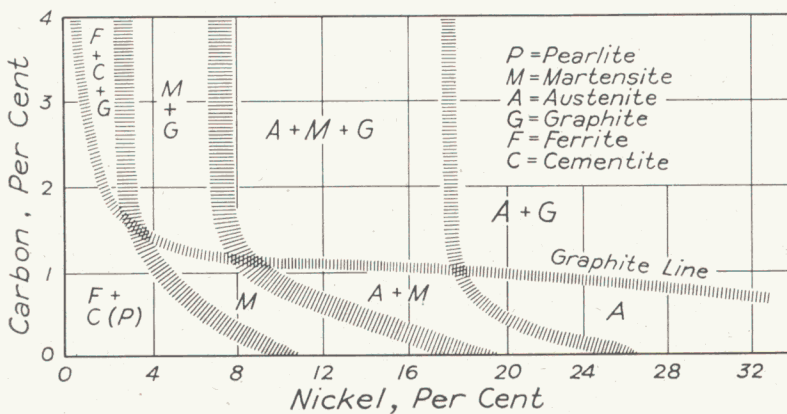


Fig. 135—The Extra-Equilibrium Constituents Developed in the Iron, Nickel, Carbon System by Slow Cooling. (*International Nickel Co.*)

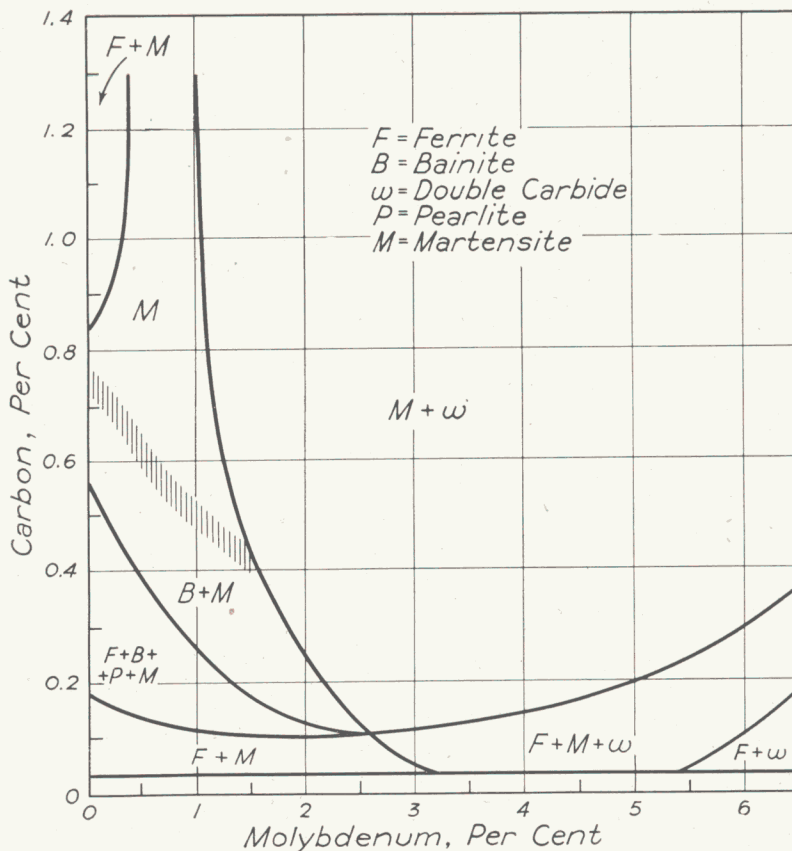


Fig. 136—The Extra-Equilibrium Constituents Developed in the Iron, Molybdenum, Carbon System by a Water Quench from 1650 Degrees Fahr. (*Parke and Herzig*).

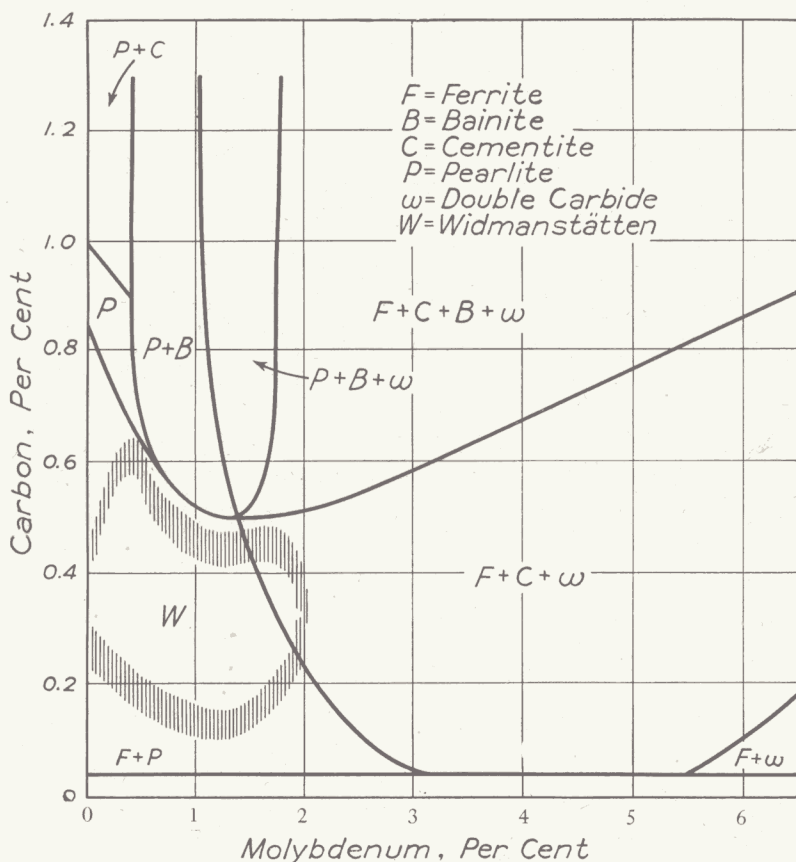


Fig. 137—The Extra-Equilibrium Constituents Developed in the Iron, Molybdenum, Carbon System by a Slow Cooling from 1650 Degrees Fahr. (Parke and Herzig).

is a part of a hardening method which must remain constant in all respects if any quantitative familiarity is to be gained with either the steel or the quenching velocity.

As a matter of collateral interest, Fig. 139 shows how the surface condition as influenced by heating practices controls even more sharply the quenching velocity in oil. The conditions parallel those of Fig. 138. The high quenching velocity of oil on a scale-free surface is shown in Fig. 140,

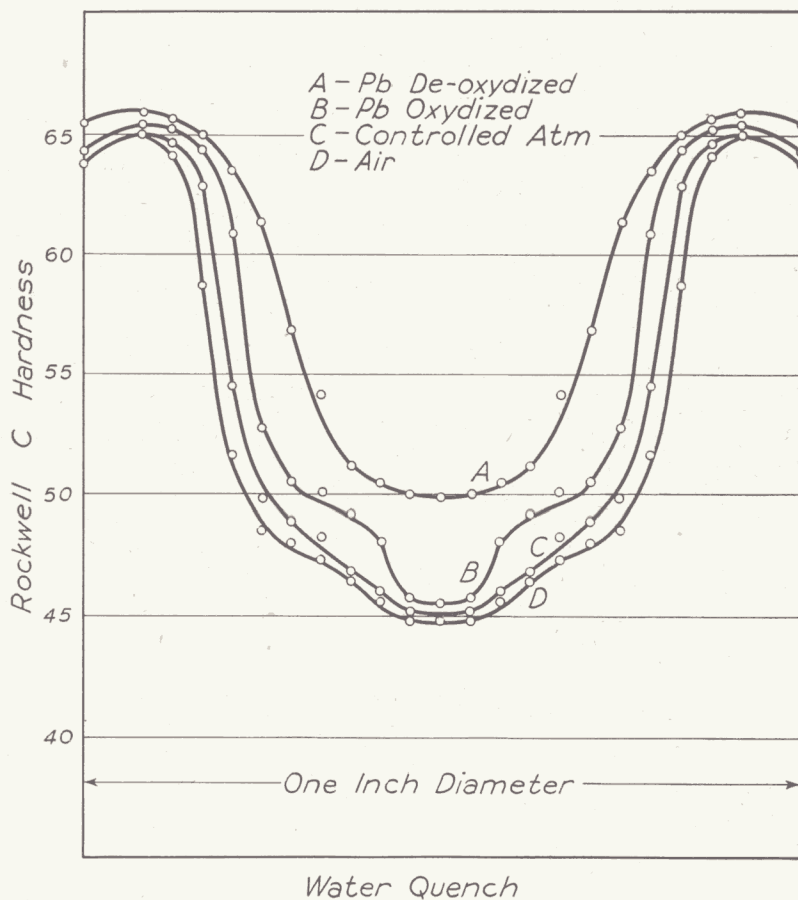


Fig. 138—Hardness Distribution in 1-Inch Rounds of a Water-Quenched Steel as Influenced by Surface Scale. Other Factors all substantially Constant. (See Footnote 43).

which reveals that a considerable proportion of martensite was able to form on the surface of a eutectoid carbon steel when the oil was in contact with practically bright, bare metal. This is, however, an unusual situation.

Significance of Heating Temperature — This would appear to be a suitable point at which to emphasize that heating temperature *in itself* has little influence upon depth

of hardening. In making this unqualified statement it is meant that, when no change of any sort occurs in the steel save that of its temperature, the quenching velocity remains



Fig. 139—Hardness Distribution in $\frac{1}{2}$ -Inch Rounds of an Oil-Quenched Steel as Influenced by Surface Scale. All Other Factors substantially Constant. (See Footnote 43).

very little changed. Actually a number of changes generally occur within the steel when it is heated to higher and higher temperature; grain size increases, homogeneity is improved, more carbide may dissolve, or more scale may form providing greater thermal insulation. One should thus bear in mind that when a steel appears to increase in hardenability as the

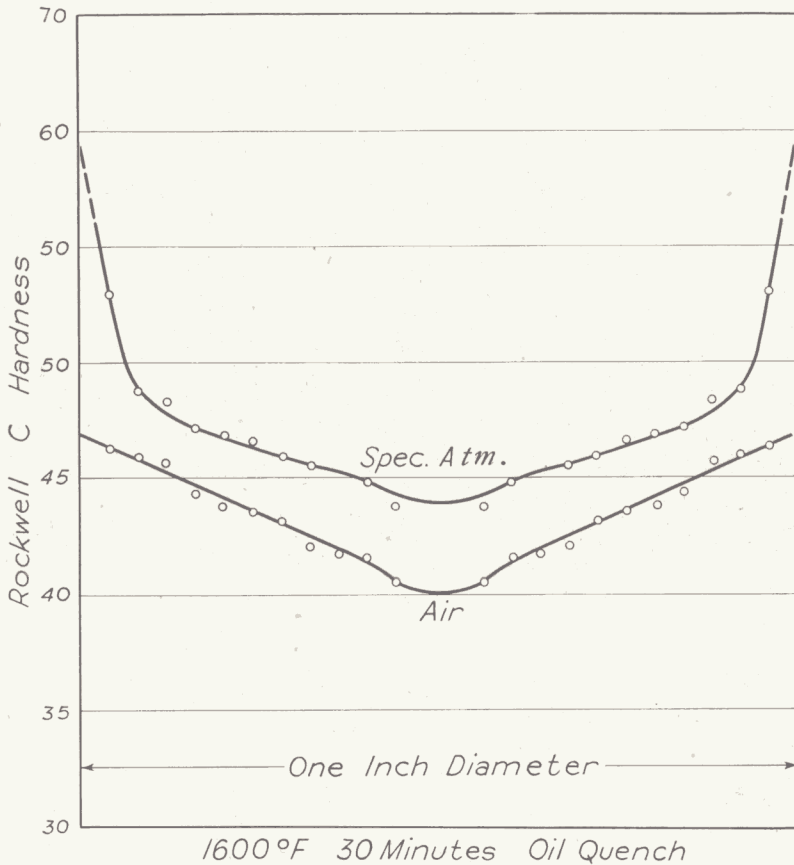


Fig. 140—Hardness Distribution in Two 1-Inch Rounds of Steel of Fig. 138, as Oil-Quenched. Note High Surface Hardness on Specimen with Scale-Free Surface. (See Footnote 43).

heating temperature (or interval) is increased, the steel is undergoing changes in structure and constituency and that these functions of heating temperature control hardenability, — not the heating temperature, *per se*.

UNIQUE STRUCTURES IN ALLOY STEELS

Thus far we have deferred any extended reference to a most important characteristic of steels carrying sufficient of

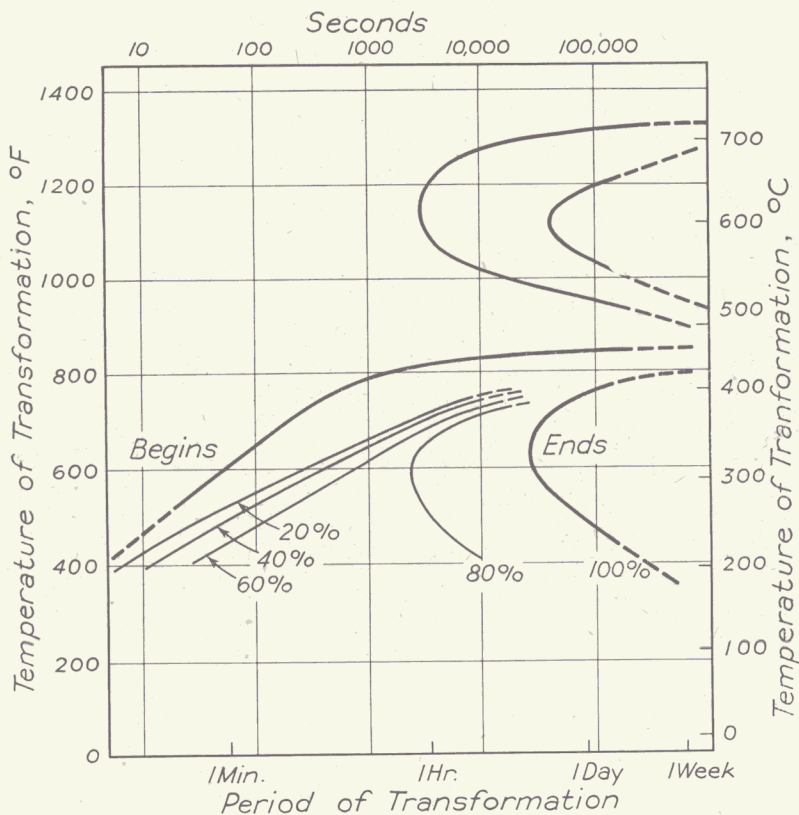


Fig. 141—The S-Curve of Transformation Velocity for a Nickel-Chromium Steel. (Davenport, Kelley, Grange and Rutherford, United States Steel Corporation Research Laboratory).

certain deep-hardening elements to impart to them a large critical size but not too great tendency to retain austenite. This has been avoided in the interest of simplicity and logical development of the subject, but the fact is that in some of the higher alloy steels it is easily possible to develop structures which thus far have only been mentioned in connection with transformation at constant temperature. Whereas in carbon and low-alloy steels the special acicular structure of moderately high hardness and great toughness requires for its formation a rapid quench, halted at intermediate temperature, it

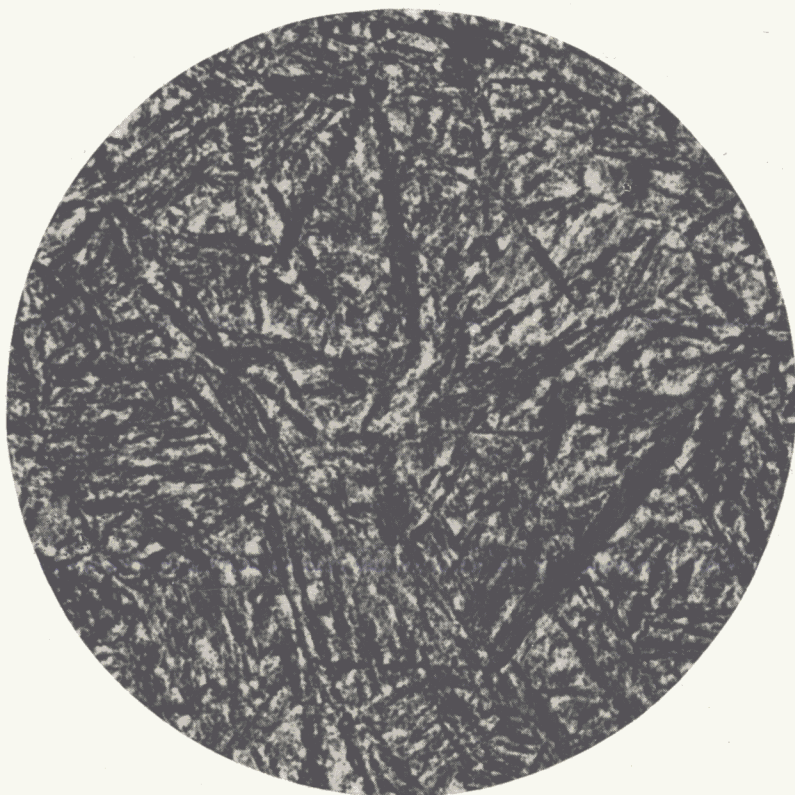


Fig. 142—The Structure of the Direct Transformation Product in a Nickel-Chromium Steel, Transformed at 650 Degrees Fahr.; Rockwell C Hardness 43. (1000 \times). (*Kelley, Grange and Rutherford, United States Steel Corporation Research Laboratory*).

may in some part be secured in certain higher alloy steels by a relatively slower cooling in a medium at ordinary temperature. When this combination of circumstances is realized some of the superior attributes of transformation at intermediate temperature are secured. Indeed, only with high alloy content and the presence of certain of the elements is there any possibility of this peculiar advantage. Its existence was taken as one of the benefits of alloy steel, but its explanation is thought to lie in some characteristics of the so-called

S-Curve of transformation velocities as lately studied.

Comparison of S-Curves — Referring to Figs. 101 and 102, it will be seen how the alloying elements retard transformation and how this is shown as a shift to the right of the S-Curve. So long as the S-Curve passes the abscissa of only 2 or 3 seconds at a temperature of about 1000 degrees, there is no possibility for the intermediate acicular structure to be formed by any cooling in a bath at ordinary temperature. The natural cooling rate cannot be so great as to escape the transformation at Ar' and then so slow as to allow transformation above some 400 degrees Fahr. in a carbon steel. But if the S-Curve were distorted sufficiently this could occur and indeed does, for certain elements distort the pattern of transformation rates in a suitable way.

Fig. 141 shows the S-Curve⁴⁴ for a 0.35 per cent carbon steel of the following composition:

Carbon	0.34
Manganese	0.22
Nickel	3.89
Chromium	1.87

Clearly a fairly slow quenching rate, possibly 30 or 40 degrees Fahr. per minute, will preclude the formation of soft pearlitic products and thus make it possible for *some* varieties of acicular structures to form. A further consideration of the S-Curve indicates that the slowest cooling above the critical rate may still make possible some transformation to acicular products other than the hardest martensite for the steel. To the extent that such transformation may occur the product is improved. Indeed cracking is unlikely and since a fair hardness results there is a gain in cooling at the minimum possible rate; provided only that the hardness is adequate.

In the course of natural, uninterrupted cooling it is not possible to secure a uniform transformation to the desired

⁴⁴Unpublished work of E. S. Davenport, S. J. Kelley, R. A. Grange and J. J. B. Rutherford of the United States Steel Corporation Research Laboratory.

acicular structure, as is effectively done in substantially constant-temperature transformation, i.e., austempering. The heat-evolution due to transformation will not precisely offset the natural loss of heat, and at last a considerable portion of the alloy steel will transform to martensite; but in part the alloy brings about the special structure. Tempering when applied will, of course, act upon the martensite and a lowering of hardness will occur. Fig. 142 shows the pure, special, acicular structure with a hardness of 43 Rockwell C as formed in the steel of Fig. 141.

Individual Alloy Effect on Special Structure—Clearly it will aid in securing a significant proportion of the special acicular structure if the transformation rate is low at higher temperatures below A_{1c} and high at intermediate ones in the range. Chromium exerts an influence in this general direction as do also the stronger carbide-forming elements. Nickel and manganese alone tend to encourage retention of austenite but in conjunction with chromium and molybdenum, for example, tend also to create steels with this special attribute in moderate cooling.

It may well be that a special field of usefulness will be filled by high-alloy steels, air-cooled in moderate section, oil-quenched in very large section, handled in the way described to produce heat-treated parts of low internal stress. When a considerable part of the steel is able to transform in a manner corresponding to that for bainite the improved toughness of the subsequently tempered structure should represent an optimum in properties from a simple, easily provided treatment.