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FOREWORD

THE PAPERS contained in this volume were presented at the first of a series of annual metallurgical conferences sponsored by the Department of Metallurgy of the School of Mineral Industries, The Pennsylvania State College. It was most appropriate that the inaugural meeting of this series should be a joint meeting of the four Pennsylvania chapters of America's most active metallurgical organization, the American Society for Metals.

Here, at the geographical center of the State, gathered members of the Philadelphia, the Pittsburgh, the York and the Lehigh Valley chapters and also, because its membership included members from northern Pennsylvania, the Southern Tier Chapter of New York.

Each of these chapters was represented by a speaker on the program and the sixth speaker was sponsored by the College. The high quality of the papers is proof of the care exercised by each chapter in the selection of the speaker to represent it. Great credit is due to the officers and committees of the several chapters for the splendid way in which they cooperated with the representatives of the College in making the meeting a success.

While four of these papers have been published since the meeting, either in the journals of the Society or elsewhere, it is felt desirable that all shall be included in this special volume as a continuing record of metallurgical conference activities.

The Pennsylvania State College welcomes meetings of this kind because they provide a medium for the exchange of ideas and are of mutual advantage to the College and to the technical interests involved.

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Department of Metallurgy.

CONTENTS

	PAGE
1. Research in the steel industry, by John Johnston	11
Discussions by: Bradley Stoughton	22
Howard Stagg	22
Francis B. Foley	23
Gilbert E. Doan	23
Closure by John Johnston	23
2. The habits and laws of decomposition of super-cooled solutions with special regard to austenite, by G. B. Upton	24
Discussions by: H. A. Schwartz	48
R. F. Mehl	49
Gilbert E. Doan	56
Closure by G. B. Upton	57
3. The P-F characteristic of steel, by B. F. Shepherd	59
Discussions by: A. E. Focke	80
H. G. Keshian	82
F. H. Allison, Jr.	82
R. Schempp	85
L. D. Hawkrigde	87
G. V. Luerssen and O. V. Greene	88
A. H. d'Arcambal	90
S. C. Spalding	92
A. C. Jones	92
Henry Wysor	93
Thomas G. Foulkes	93
Gilbert E. Doan	93
J. A. Mathews	94
Closure by B. F. Shepherd	94
4. Steels: Their selection and application in general shop practice, by Frank J. Allen	99
Discussions by: H. B. Northrup	111
5. Wrought aluminum alloys for structural applications, by E. H. Dix, Jr.	113
Discussions by: Lloyd M. Morris	120
Norman I. Stotz	120
J. W. Foster	121
Horace C. Knerr	121
Closure by E. H. Dix, Jr.	125
6. The effect of gases on ferrous materials at high temperatures and high pressures, by H. L. Maxwell (Abstract only).	128
Discussions by: Franklin J. Allison	129
T. Holland Nelson	130
J. H. Stoll	134

ILLUSTRATIONS

	PAGE
The Habits and Laws of Decomposition of Supercooled Solutions, with Special Regard to Austenite	24
Figure 1—Stoughton and Greiner Iron-Silicon Diagram	26
2—Upton Diagram of First Part of Iron-Silicon System	27
3, 4—Sato's Data of Supercooling and Superheating, Hypo- eutectoids	28, 29
5—Sato's Data; Critical Temperature, Superheating and Supercooling of Pearlite	30
6—Fe-Fe ₃ C Equilibrium, Drawn by Upton from Sato's Data	30
7—Composition and Precipitation Temperature of Decom- position Products of Austenite (Pearlite and Marten- site) at Various Cooling Rates, and After Various Stabilizations	31
8—Data of Ruer and Klesper on Supercooling in "Pure" Iron	32
9—Relations of Stable Fe-G and Unstable Fe-Fe ₃ C Systems at Eutectoids (a) and Eutectics (b)	34
10—Decomposition of Austenite at Constant Temperature, at Various Temperature Levels under the A ₁ Critical Temperature (Davenport and Bain with Additions by Author)	36
11—Form of Time History of Pearlite Precipitation from Supercooled Austenite, at all Temperatures Where It Occurs	37
12—General Effect of Temperature on Reaction Rates	40
13—Form of the Function Ψ (T _c -T) the Effect of Distance Below Its Equilibrium Critical Temperature on Time Rate of a Reaction	41
14—Davenport and Bain's Figure of Hardness of Completed Transformation Products of Constant Temperature Transformations of Austenite, with the Author's Mark- ing of Martensite Critical Temperatures of Each Steel Added to Their Figure	43
15—Dilatometer Curves of Transformation of Austenite at Constant Temperatures	44
16—Solubility Relationships as Affected by Crystal Size	45
Written Discussion—R. F. Mehl—Figure 1	55

The P-F Characteristic of Steel	59
Figure 1—Set of Martensitic Fracture Grain Standards Showing Protective Covers Removed from Specimens 2 to 10	60
2—Photomicrographs of the Structure of Fracture Grain Standards Shown in Figure 1	62
3—Showing Method of Quenching Specimens in a Standard Vertical Jig	63
4—Gage for Measuring P-Value	64
5—P-F Test Section Showing the Uniformity of Distribution of Hardness Along the Length	65
6—Comparison of Rate of Heating a $\frac{3}{4}$ -inch Round Specimen of Electric Furnace 1.05 per cent Carbon Tool Steel in an Electric Muffle Furnace and Lead Pot	65
7—Influence of Structure on the Critical Point of $\frac{3}{4}$ -inch Round 1.05 per cent Carbon Tool Steel	66
8—A Standard $\frac{3}{4}$ -inch Slab Test Piece and Two Test Pieces from Different Heats Showing Varying Degrees of Dumb-bell Fracture	69
9—Polished and Etched Specimens Showing the Uniformity of Hardenability Through the Entire Heat of Steel (Shepherd Hardenability Test)	71
10—Etched Sections of Half of the Cross Sections of Billets from Steel B-332.	72
11—Bars Rolled from Billets Shown in Figure 10	73
12—Influence of Inclusions on the P-F Characteristics of Carbon Tool Steel	74
13—Exploration of Heat 31196, Specification 1 for Uniformity of Penetration-Fracture Characteristic	75
14—Exploration of Heat 1D006; Specification 332 for Uniformity of Penetration-Fracture Characteristic	76
15, 16—Rockwell Hardness Survey of Steel X30864 Hardened as Shown	77
17, 18—Rockwell Hardness Survey of Steel X29983 Hardened as Shown	78
19—Vickers Hardness Survey of $\frac{3}{4}$ -inch Round P-F Test Pieces	79

RESEARCH IN THE STEEL INDUSTRY

by JOHN JOHNSTON

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AS ONE who is by training and experience a physical chemist rather than a metallurgist, I may have a point of view somewhat different from that of those who are more directly concerned with metallurgical operations, a somewhat different philosophy of the broad question of the making of steels best fitted for the purpose which they are to serve. It is my aim to outline this general philosophy, rather than to discuss specific details, and to indicate my present views as to what systematic investigation can accomplish toward further improvement in control over the art of steel making, and toward providing in uniform quality that steel which is best fitted for each particular use, with reference, particularly, to the more and more exacting conditions which the engineer is asking steel to meet. It is a large subject to discuss in a short time; it is obvious, therefore, that many of its aspects must be omitted, and that only a few examples can be given of the kind of precise information requisite as a sound basis for further progress. These examples have been chosen as being somewhat less familiar to you who are, I presume, users rather than makers of steel.

Perhaps it would be well to begin by considering the debt which the art of steel making already owes to scientific investigation. In such an ancient art it is not easy to state this debt precisely, because, until recently, the scientific investigations which have in fact benefitted the steel industry have been carried on entirely outside that industry, largely indeed without thought of benefiting any industry. But there can be no question that without the aid of science the condition and magnitude of the steel industry would have been very different. For the great expansion of the steel industry since the beginning of the present century has been brought about largely by the growth of other industries, such as the electrical and automobile industries, which followed upon the application to them of science based ultimately upon the work of men such as Faraday and Clerk Maxwell. Conversely, these industries could not have developed as they have without the technological advances already realized in the making and finishing of steel on a large scale.

A familiar example is that without alloy steels there could be no high-speed tools, with the incalculable advantages which they brought with them; but these alloy steels could hardly have been developed until the alloying elements themselves had been discovered and their preparation and properties had been studied. Indeed, a century passed before much application had been made of this knowledge, most of the elements used in steel having been described before 1800; and in the proper application of alloying elements much more remains to be done than has yet been achieved. Again, the making of steel is now continuously controlled by repeated chemical analyses, by use of methods which were, in the first place, invented without reference to industrial use and developed over decades before any one thought of using them regularly to control the gross composition of steel. But these methods give us information only as to the relative amounts of the several constituents determined by them; they tell us nothing as to the way in which those constituents are combined with one another. In other words, they tell us the kinds and quantities of our primary building materials but nothing as to the architecture of the building. But just as the architecture of a building determines its usefulness, so does the architectural structure of a steel determine its practical usefulness; consequently, something more than analytical chemistry was needed.

Knowledge of the mode of combination in steel of the constituent elements has grown rapidly, especially during the last 20 years, with the systematic use of other methods of investigation—for instance, the microscope, the X-ray spectrograph, the dilatometer—combined with increasing attention to important factors such as the temperature, and duration, of the treatment of the metal. The proper interpretation of this information derives from an abstruse paper published nearly sixty years ago by Willard Gibbs, the greatest imaginative scientist yet produced by America, and one of the greatest of all scientists. Gibbs, however, was so far ahead of his time that many years elapsed before chemists discovered the profound significance of his work, which has in the last quarter century largely revolutionized the science of chemistry. With the application of Gibbs's thermodynamic principles in many laboratories, by many men, in many countries, it became possible to correlate, hence to control, many phenomena which previously had seemed to be separate and even mysterious; and many methods, based on this work, are now used every day in the steel industry, frequently with little appreciation of the difficult road traveled in perfecting the tools and methods of investigation and in learning to interpret the observations correctly.

This development was possible only by collateral development of other knowledge and other technical advances based upon research in pure science; for instance, methods of accurately measuring and controlling temperature, of securing suitable refractory furnace linings; the whole technique of the microscope, of

the preparation of specimens for such examination, and of the identification by this means of the many different kinds of microscopic structure, helpful or harmful, which may occur. Indeed, a very interesting book could be written which would trace the development of the technical side of steel making back to the sources; but the ramifications into physical science would prove to be so many and various that the writing of such a book would be long and difficult.

I recount this to remind you that the steel industry has made considerably greater application of the results of scientific research than it is commonly given credit for; though it must be admitted that there was a large time lag between the discovery of this knowledge and its utilization, and that many possibilities had not even been considered. This, indeed, was true of nearly all industries until very recent years, when those most alert to improvements in their tools and processes discovered that they could lessen this time lag from years to months by having their own research group whose duty is to investigate *systematically* the possibilities of improvement in process and product by the utilization of new knowledge and the application of new scientific tools. This policy, properly carried out, proved so successful that there are now few major industries which have not adopted it in some form, so much so that it has by some been asserted to be a major cause of the depression from which we are now slowly emerging. The rapid technical development of other industries brought with it demands for better metals, for metals to meet new and far more exacting specifications, demands which at the time the art was unable to meet satisfactorily.

The enormous expansion of the steel industry in this country has come about as a direct result of the cheapening of steel made possible by mass production. This has involved wonderful development work, as exemplified in the building and operation of larger and larger furnaces, of larger units generally, the replacement of human labor by machines with corresponding greatly increased use of power, nevertheless, with a lessening of the total fuel used per ton of finished product. These achievements are impressive and of the highest importance economically; without them steel would have cost so much more that its use must have been far less extensive than it has been. For it is the amount not merely of the steel in your car, but of the steel in all of the large number of machines and factories taking part in the production of the finished car, which enable your car to be produced on such a large scale, hence so cheaply, as it is. These achievements in organization and production have gone so far that no further appreciable lessening of cost of ingots and billets can be looked for from further mechanization. Indeed, a large fraction of the present cost of semi-finished steel is for the raw materials delivered at the plant, a large, and in recent years increasing, fraction of this large item of cost being for transportation of those materials

to the steel plant; and this part of the cost of steel seems likely to continue to rise more than can be compensated by any possible further improved efficiency in the use of the raw materials. The possibility of improving the economy of steel to the consumer is therefore largely a matter of improving its uniformity of quality, of fitting steels better for each of the multifarious uses, rather than of any direct lessening of its cost of production.

The steel produced has been, and is, entirely satisfactory for a great many of the purposes to which it is put; but with the rise of the automobile industry, a period which began approximately with the World War, there arose a continually increasing demand for steel of higher, and more uniform, quality at a low price. This insistent demand raises many problems, problems which are not likely to be solved except by thorough systematic investigation, in part of things which have been quite generally supposed to be definitely known and established on a firm basis of fact. These problems fall into two apparently rather distinct categories; yet they are so closely intertwined that they cannot in practice be separated:

(1) Development of further and more precise knowledge of the factors which determine the useful properties of a steel, as a means of selecting consciously that steel which is best fitted for any particular use; in other words, the manifold problems involved in finding out what should be made.

(2) Further improvement in control of the processes as a means of securing still better uniformity of any given kind of steel; in other words, the problems involved in actually making just what is really wanted.

These large questions obviously involve a study of the chemistry, if you will the physical chemistry and physics, of the processes of steel-making and of the steels produced in all the various ways. The solution of these manifold problems and the practical application of this knowledge require the use of tools and modes of control more sensitive and more reliable than the operator's unaided senses, no matter how skilled he may be. These tools and methods must be, and will be, invented and developed; they range from the accurate measurement and control of the highest temperatures to the provision of better refractories; from rapid methods of analysis for dominant factors in slags and in liquid steel to rapid tests of quality on the semi-finished steel; from simple chemistry and physics to recondite applications of electron behavior; from simple proportion to the use of modern statistical methods which enable one to use the steel plant as the laboratory.

Steel making is an ancient and wonderful art, developed through a long experience; perhaps because of this, it is far from easy to discover the precise facts underlying the processes used in the art. A question elicits varied responses which differ in emphasis and may even seem to be contradictory. For example, there is a

wide difference of opinion among practical men on questions such as: In how far is the quality of a finished steel of fixed chemical specifications influenced by apparently small differences in raw materials, or in furnace practice, or in deoxidation practice? Should steel be poured "hot" or "cold"? What kind, size, and shape of mold should be used? What is the best rolling practice? and so on. Indeed, one may venture the remark that no two steel makers produce the same steel in *precisely* the same way, just as no two artists produce the same effect in *precisely* the same way. Yet it is far from certain that a combination of what might seem to be the best features of all the present methods and artifices would yield even as good a product as the skilled operator now makes.

Altogether, then, there had been, until recent years, but little systematic thorough investigation of many of the really fundamental problems involved in the proper making and use of steels; and many of the studies which had been made had led only to somewhat indefinite, even ambiguous, conclusions. This pointed to the necessity of making rather fundamental investigations, to secure a firm basis upon which one could, with confidence, start building, without being too much concerned with achieving immediate improvements in manufacturing practice. But what may one take as reliably known? The metallurgical books seem to assume that all is known, and generally state the matter in simpler terms than correspond to the facts. The chemical books give, with respect to the element iron, its reactions, and compounds, so little concordant information that they are not very helpful. One is therefore constrained to rely largely on one's judgment and general experience in deciding where and how to start, and as to what may safely be accepted as a firm foundation on which to build and what is still doubtful or unknown.

We were, however, convinced that certain principles—which may be designated as the principles of physical chemistry—would be found to apply to this more complex case just as they are known by experience to apply to simpler chemical cases, if we had the wit to make the appropriate measurements and to apply these principles properly. In other words, the seeming anomalies and mysteries result from the complexity of the system, from the large number of variable factors, and not from the intervention of some new and strange principle. And we are now entirely persuaded that this is so.

A second general guiding principle, which so far has not failed us, is that the physical and mechanical properties of any ordinary or low-alloy steel—say, any steel containing more than 95 per cent of the element iron—depend upon the final intimate architecture of the metal, and not upon the precise way in which this detailed structure came about. A particular structure may happen to have been built up by some special deoxidation practice, by

an alloy addition, by a mode of working, by a specific heat treatment, or by some combination of these (or other) factors; we have therefore to discover which combination of the possible methods yields most easily, that is, at the lowest overall cost, the properties desired in the finished product.

Consideration of this matter brings us back directly to a very fundamental question, Wherein precisely does iron (or steel, if you will) differ from the other common metals? To what peculiarity of the element iron, apart from its abundance and cheapness, is due its supreme position as the most useful metal? There is, in my mind, no doubt as to the answer: Of the common metals iron is the only one (except for tin, which is so different that it need not concern us here) which occurs in two distinct crystalline forms, and it so happens that these two forms have a very different solvent power for other atoms, for carbon atoms especially. From this arises directly the possibility of producing very different results by varying the mode of heat treatment of steel; otherwise, many things now commonly done would be impossible, or at least impracticable on anything like the present scale. For the success of all of these heat treatments depends ultimately on the facts that carbon is much less soluble in α -iron (ferrite, the form stable at low temperatures) than in γ -iron (austenite, the high-temperature form), and that the mode in which the carbon appears in the finished steel depends upon the way in which we cause it to precipitate, upon the way, therefore, in which we cause the crystals of γ -iron to transform to α -iron. Moreover, the wide variation possible in the temperature range within which the crystals transform depends upon the circumstance that in steels the transformation $\gamma \rightarrow \alpha$ proceeds at a relatively slow rate; and this slow rate (which is far from universal in transformations of this type) enables us to make the carbon precipitate at any temperature within a wide range.

Incidentally, it may be mentioned that within recent years a number of strong alloys, notably of aluminum, have been developed, which are based upon a precipitation of another element (or compound) brought about by the change with temperature of its solubility in a single crystalline form of the principal metal. The proper use of this general method of increasing strength and hardness of a metal demands a degree of refinement in control of the operations which has become feasible only within recent years, whereas the hardening and tempering of steel have been practiced successfully for centuries.

To return now to the crystal transformation of iron in steel in the so-called critical-temperature range, a somewhat misleading term which I would like to see go out of use. A steel when heated begins to transform at a temperature very near to the real initial equilibrium temperature, no matter what the rate of heating, and the transformation is complete before the temperature of the steel actually transforming has risen more than a few degrees above

the final equilibrium temperature. When cooled, however, it does not transform in the equilibrium-temperature range unless it is cooling very slowly. The more rapid the cooling, the lower is the actual transformation temperature; with very rapid cooling, as in quenching, the steel behaves differently, the resultant product being analogous to (and possibly, identical with) a supersaturated solution, or better, to a glass which on subsequent reheating (tempering) partially or completely devitrifies—that is, transforms to a more stable state in accordance with the inherent rate of reaction at the temperature.

At every actual temperature of transformation there is, *for a given steel*, a definite speed of transformation, and correspondingly, a definite final structure; and both the speed and the resultant structure change very markedly with change in the temperature at which the transformation actually occurs. At temperatures just below the equilibrium transformation temperature (that is, the A_c point) the product of transformation is a coarse pearlite, a banded, or lamellar, structure; at lower temperatures, the pearlite produced is finer and finer, and finally so fine that it is unresolvable under the microscope. This is most directly shown by causing small pieces of the steel to transform in a metal bath, held at one of a series of constant temperatures; but it can also be brought about by varying the rate of cooling prior to the onset of the transformation. This general phenomenon is entirely analogous to many well-known examples of precipitation and crystallization of barium sulphate, for instance, which can be produced in crystals ranging in size from several millimeters to sub-microscopic, and even ultramicroscopic, merely by changing the degree of virtual supersaturation of the solution within which it is precipitated.

Alloying elements alter the transformation temperature, in some cases lowering it, in others raising it; more significantly, they slow it up, in some cases tremendously. Thus a given cooling rate results in a lower actual transformation temperature, and in the structure corresponding to this lower temperature of precipitation of the carbide; that is, it results in a finer pearlite and consequently in a harder and stronger steel. This we believe to be the outstanding effect of an alloying element, though, of course, some of them are, or are also, chemical reagents in the finishing stages of the steel-making process, and by so reacting modify the *real* composition of the steel.

I stress *real* composition, because what has been accepted as the composition of a steel—its so-called chemistry, as determined by the classical methods of analytical chemistry—does not in itself tell the whole story of the behavior to be expected from the steel. For its actual behavior is determined not only by the proportions of the elements usually analyzed for, and of the alloying elements added intentionally, but also by the presence of very small proportions of elements such as oxygen, nitrogen, and possibly hy-

drogen; and in these cases especially it is less a matter of the proportion than of the mode of combination in which the element is present and its mode of distribution in the metal. For instance, the influence of a small proportion, say 0.01 per cent, of oxygen, is very different according as it is present wholly in the form of mineral particles, such as alumina or silicate inclusions, or in solid solution in the crystals of the metal itself. This leads me to state our present belief that in general the useful properties of a steel are affected much less by the mineral particles visible under the microscope—the so-called non-metallic inclusions—than by those which with present technique cannot be seen at all. This is another example of the large effect which may be produced by the presence of a very small percentage of the proper material, properly distributed; as an analogy we may think of a masonry structure, the strength of which depends on what was used as cement and how it was laid on, even though the total weight of the cement is but a very small fraction of the total weight of the structure.

The circumstance that two steels of identical nominal composition may differ markedly in their response to certain treatments—a circumstance which has been generally admitted only within the last two or three years—suggests that many of the specifications for steels lay stress on some things which are of far less significance to the user than other points which are not even mentioned in the specifications. Undue emphasis has been laid on the ordinary chemical specifications; and this has brought about the situation that the steel maker has to make steel to five, or even ten, times as many different chemical analyses as are really necessary. This excessive number of chemical specifications implies that the user not only ultimately pays somewhat more for his steel but also that in general he gets a metal of less uniform quality than he would get by buying a steel which the steel mill could make in a large number of successive heats. After all, the user is not really concerned with the gross chemical composition of the steel; what does concern him is its intimate structure, no matter how it may have been secured, and how it fulfils his purpose. In specifications I should like to see relatively more emphasis put on this aspect of the matter and less on the ordinary chemical analysis.

A kindred point is that, in our opinion, undue stress is laid in specifications upon the difference between bessemer steel and open-hearth steel. It is true that there are differences between them, as usually made; that bessemer steel is more suitable for some purposes, open-hearth steel better fitted for others. But these differences must be attributed to differences in real composition, for it is to me inconceivable that steels of *completely identical* real composition should not have completely identical properties, no matter from what materials or by what process they were made. Indeed, the actual result depends upon the conditions un-

der which the process was carried out, particularly in the finishing stages, perhaps more than upon the type of process in itself. For the reactions are essentially identical in both processes; in the bessemer we oxidize directly by blowing atmospheric oxygen through the liquid metal, whereas in the open-hearth furnace we oxidize the liquid metal by means of the iron oxide in the slag in contact with it. Thus the two processes differ only in the means by which the reactions are brought about, and in practice to some extent because in general they are used to make different types of steel. Similar remarks apply to steels made in the electric furnace, which are not necessarily superior merely because they were made in an electric furnace.

The character of a steel of a given chemical analysis is influenced by the way in which it is finished, in particular by the real composition of the liquid steel and of the slag then in contact with it, and by the mode of "deoxidation." It is not yet certain whether this so-called deoxidation is merely a partial or complete precipitation of oxygen from the steel, or involves other simultaneous reactions, as with nitrogen or sulphur. But oxygen appears to be the dominant factor, though just what happens may be influenced by the presence of these other non-metallic elements. Nor do we want an absolutely pure steel, for such steel would be largely useless for most purposes because it would be coarse-grained and remain so in spite of repeated heat-treatments. What is wanted is a steel with just the right amount and kind of non-metallic matter, properly distributed through the steel; I was going to say dirt, but dirt is defined as matter in the wrong place, whereas here we want matter in the right place. You can see that all of this makes it difficult to reach a definitive solution of this whole question; but very satisfactory progress in this direction is being made now when the real question is being faced by the steel maker, who is going to furnish to the engineer the steel in uniform quality which will best meet his more and more exacting requirements as he proceeds to employ higher and higher unit stresses, higher temperatures and pressures, and so forth.

You are not to infer that this difficulty of securing uniformly the best real composition of the liquid steel is the sole obstacle in the way of producing steel of absolutely uniform highest quality, for perfectly good liquid steel in the ladle may be spoiled by improper subsequent treatment—by improper pouring, heating, rolling, or finishing. But it is true that a steel which in the ingot was not right for a particular use, cannot, in general, be made right by subsequent juggling.

From these remarks, it will, I think, be clear to you that, whereas to make a steel within close limits with respect to the ordinary chemical analysis is relatively easy, to make a steel of absolutely uniform high quality, heat after heat, is far from an easy matter, involving as it does some very elusive factors. Moreover, some of the factors encountered in practice are opposed in effect, and

therefore there must be some sort of compromise between them. For instance, to minimize the phosphorus content necessitates a temporarily high oxygen content which must somehow be removed later. Indeed, I am tempted to wonder if the presumed adverse influence of phosphorus is not in part because the steel quality may have been spoiled in endeavoring to bring the phosphorus content down beyond where it is actually necessary to bring it. For we are persuaded that phosphorus, within reasonable limits, is a hardening element in all respects analogous to carbon, and should be taken into account in comparing the properties of different steels. Indeed, a statistical comparison of a very large number of steels shows that the hardening or strengthening influence of one point of phosphorus is equivalent to that of about 3.5 points of carbon. On this basis a steel with 0.08 carbon and 0.02 phosphorus is equivalent to a steel carrying 0.15 carbon with a negligible percentage of phosphorus. When compared on this basis, some of the differences between ordinary bessemer steel and open-hearth steel disappear.

Incidentally, it is a fortunate circumstance that at ordinary steel-making temperatures most of the elements we wish to remove oxidize more rapidly than does iron. But these relative rates can be changed, for instance, by change of temperature; for example, chromium usually oxidizes out ahead of carbon, but there is good evidence indicating that at higher temperature the relative rates reverse and that then carbon will be removed by oxidation ahead of chromium. But to do this in practice involves improvement in refractories, which is a large and difficult question in itself; the difficulty is perhaps less in getting something to stand the high temperature than in finding something which will withstand the solvent action of the iron oxide in the slag and the apparently unavoidable rapid temperature changes. In fact, there is no steel-making problem which does not immediately branch out into a great number of problems, so many, so diverse, in part so difficult experimentally, that they will not be solved satisfactorily all at once or for a long time to come. What has been accomplished recently is bringing about almost a revolution in the methods of making quality steel and in our knowledge of its intimate architecture; for instance, as to just how seemingly slight variations in the make-up of the metal may bring about significant differences in this architecture, hence in the useful properties of the steel. Incidentally, it seems to us more logical to draw a distinction not between alloy steels and plain carbon steels as is usually done, but between quality steels—which may or may not contain special alloying elements—and ordinary steels.

There is an important implication of the fact that the quality of the finished steel is determined not alone by its composition deduced from the ordinary chemical analysis, but also—and in some cases, more significantly—by other factors, such as grain size which affects the response of the steel to heat-treatment, and

the precise mode of heat-treatment to which the steel had in fact been subjected, whether by intention or not. It has been usual to attribute differences in the measured mechanical properties of steels to differences in given composition, as determined by ordinary analysis, as if this were the only significant variable; with the consequence that there are great variations in the value of any of the mechanical properties assigned by different authors to steel of any given nominal composition. As a further consequence, there is still considerable uncertainty as to the precise influence of a variation in the relative proportion of the several elements, whether added intentionally to make so-called alloy steels or not, upon any of the mechanical properties of steel; and this difference of opinion is doubtless the main source of the excessive number of separate chemical specifications which the steel maker is called upon to meet. The clarification of this whole matter now in progress will, in time, put the engineer in possession of more precise and more reliable data on the properties of the quality steels available for his use; and this, with the greater uniformity of the steel, will enable him to design more precisely, with a still smaller so-called factor of safety and a further lessening of dead-weight of his structure or equipment.

In conclusion I wish to refer briefly to the degree of significance of mechanical tests and to suggest that the mechanical engineer look more closely at them, preferably, perhaps, in collaboration with a metallurgist familiar with the previous history of the specimens tested. For it looks as if some of these mechanical tests, as now carried out, may not have the significance commonly ascribed to them, except as purely conventional engineering comparisons. The usual interpretation of the results implies tacitly that the material was strictly homogeneous and initially stress-free, and that its architecture remained essentially unaltered by the deformation imposed by the test; but no one of these assumptions is rigorously true. This matters little so long as a large factor of safety was included in the computations, but must be kept in mind as the metal is used under more and more exacting conditions. Moreover, all of these tests measure some mixture of cohesive strength and resistance to slip, but not in the same proportion for different metals, or even for the same metal at different temperatures of test. Even the so-called elastic limit is observed at lower and lower stress as the sensitivity of the means of observation is increased. It would seem to me preferable, therefore, to express the results of these mechanical tests in terms of the stress required to produce one or more specified amounts of permanent deformation—that is, directly in terms of the quantities observed and required in design—rather than as some kind of modulus which would be valid only if the stress-strain curve were strictly linear over the whole range of stresses considered.

In this connection I would like to remind you that in all cases a

service test is the final criterion, and that the results of no approximate or indirect test can be safely accepted as a reliable prediction of what will happen in service until they have been definitely correlated with the results of actual service tests. This is especially true of all accelerated tests of resistance to corrosion of one kind or another, or even of resistance to fatigue or to creep at high temperature; for in the very acceleration of the process we introduce a change in the intensity of some factor which may change some essential condition and so cause the test to be misleading as an indication of what will happen in service over a long period.

The foregoing is a very general reconnaissance of a large region of which only the outstanding features are yet known; its detailed mapping will require the work of many explorers and frontiersmen. I have tried to indicate that whereas much remains to be done before steel production is thoroughly on a scientific basis, much has been accomplished toward the ideal of controlling the processes so that they yield a uniform product with the best combination of qualities desired for any specific purpose. A good deal of the necessary fundamental information is still lacking, but this lack is now more generally appreciated, and the information will be forthcoming. The old experimental tools and technique have not all been used to their limits; new ones will be developed. The requisite investigations are not easy, but are so interesting and will yield such valuable results to the industry, that they will attract a large number of men; so that progress in real knowledge of the properties and capabilities of steels is going to be more rapid than ever before.

DISCUSSION

O. E. HARDER, *Chairman*

Assistant Director, Battelle Memorial Institute, Columbus, Ohio

PROFESSOR BRADLEY STOUGHTON, (*Department of Metallurgy, Lehigh University*): I agree with Dr. Johnston that the Bessemer and open hearth products are equally good if phosphorus and sulphur are the same, but do not see how Bessemer steel can be made equal to open hearth steel at a competitive price. There is no question of the value of research. In England metallurgy has, in the past, been governed largely by tradition. In Germany it has been governed by research. The results, in the recovery of the steel industry in these countries since the world war, have largely shown the advantages of the German method. Even though handicapped for some time by the loss of large iron ore deposits to the French, their come-back in the fuel and iron industries has been remarkable.

HOWARD STAGG, (*Halcomb Steel Company, Syracuse, New York*):

I am very glad to have had the opportunity of listening to this paper. Dr. Johnston has pointed out some of the few things with which we are struggling. The things not pointed out are appalling. The industry as a whole appreciates how much we do not know and is making an "honest-to-God" effort to solve the problems. Professor Stoughton's criticism is right. There is now much greater effort to "hit things where they grow."

FRANCIS B. FOLEY, (*The Midvale Company, Nicetown, Philadelphia, Pa.*): Dr. Johnston's talk was very stimulating. I am not greatly concerned in statistical research of the kind mentioned by Professor Stoughton, but I am much concerned about the kind advocated by Dr. Johnston. Men in the plant with accumulating experience are best able to gather such statistics and many troubles could be eliminated by attention to them. Much of the so-called "research" of the steel men was, for many years, of the "cut and try" kind. A bright spot in recent years has been the study of the physical chemistry of steel melting and refining by Dr. Hertzy and his associates. This has brought together scientific and practical men in an effective way. It was done largely under the guidance of men in the industry and will lead to new things.

PROFESSOR GILBERT E. DOAN, (*Department of Metallurgy, Lehigh University*): Why do you specify the importance of a change of solubility accompanying allotropic change? Is not the change in solubility the main fact? In aluminum, for example, we have not allotropy but do have large changes in solubility. Much of the advantage of iron over other metals lies in its relative cheapness.

DR. JOHNSTON: When I said the same analysis I meant approximately the same composition. The Bessemer process is more expensive now but has large possibilities. In regard to Bessemer steel I was not touching the economic question but felt that it could be made as cheaply and as uniform as open hearth steel.

I suspect that change in solubility can be very effective, but the sudden change at the allotropic transformation permits greater change and greater variation in the speed of transformation. If copper and aluminum had transitions we would be able to do much more with them.

THE HABITS AND LAWS OF DECOMPOSITION OF SUPERCOOLED SOLUTIONS, WITH SPECIAL REGARD TO AUSTENITE

by G. B. UPTON

Professor of Experimental Engineering, Cornell University, Ithaca, N. Y.

THE whole theory of heat treatment of metals is closely tied with what austenite—the solid solution of carbon in gamma iron—does or might do. We may by analogy study even liquid solutions; and we must study the fields bordering the austenite field. In this bordering region we offer some suggestions that may be novel.

For some time it has been fashionable to deny that alpha, beta and delta forms of pure or nearly pure iron were separate allotropic forms. That is largely a matter of definition; and if the statement of space lattice form and dimensions is accepted as constituting a definition of allotropy, the three are identical. But if the older definition of allotropy stands, that one substance in one state of aggregation (liquid, solid) may have forms differing in physical properties, such different forms being allotropes, then alpha is different from beta and delta, though the latter are identical. Study of any of the equilibrium diagrams of iron alloys having a “gamma-loop,” such as that of iron-silicon, will confirm the latter point; the beta and delta fields are one and the same.

Now not only has the old definition of allotropy been challenged, but that of a phase also in an alloy system. A phase is now being defined as a space lattice,¹ a system of sites at which atoms are placed in crystals, further characterized by a ratio between atoms and free electrons in the “unit cell” of the crystal. In this definition there remains no difference between beta and beta prime solid solutions in brasses, near 50 per cent zinc; nor between, we suspect, alpha and beta iron. In both these cases, and similar ones (for they are not few) we maintain that the old definition of physical chemistry, used for the last 50 years, must prevail over the recent suggestion. The old definition of a phase was that it was a mass or entity physically and chemically homogeneous within itself, and different physically and/or chemically from its neighbors. Under this definition alpha and beta iron are

separate phases, entitled to their own fields on the equilibrium diagrams; so are beta and beta prime solutions in the brasses separate phases. However beta and delta solutions in iron alloys are not different phases, but one and the same, as a look at the iron-silicon diagram, for example, will show.

The difference between beta and beta prime solid solutions in the brasses, or between alpha and beta iron, is one of “order” or “disorder” in the arrangement of atoms in the space lattice, or of electrons within or surrounding the atoms. In the brasses the high temperature beta solid solution has the copper and zinc atoms, of about equal numbers, arranged at random in the sites of a body-centered cubic space lattice; the low temperature beta prime solid solution has the same lattice, but the atoms are regularly arranged, in a way that may be described as copper atoms at corners of the cube, zinc atoms at the centers. The transition from one to the other arrangement of atoms, the random and the regular (“disordered” and “ordered”), can be studied mathematically, and described by the intersection of two thermodynamic curves when the arrangement is regular, one curve swinging with regard to the other as temperature changes; but at a certain critical temperature the curves suddenly cease to intersect after having just previously had a common tangency. The onset of the regular arrangement from the random, or vice versa, is thus approached continuously from either side; but there is a discontinuity at the transition. To me the mathematical treatment of the relation of the beta and beta prime phases in brasses constitutes not a showing that they are one and the same, but a very pretty proof that they are different, and a neat showing of how and why they are different.

The change from alpha to beta iron is usually thought of in connection with magnetism. This can be treated exactly as the “order-disorder” of atomic arrangement is analyzed in the beta-beta prime case of the brasses, though now we deal with atom and electron relationships, as is shown by the fact that atoms of iron take magnetism with them into many chemical compounds. Along with the magnetic change many other physical properties, tied with the atom-electron relationships, are different in alpha and beta iron.

Proof that alpha and beta iron may well, or must, be treated as separate phases, because so doing greatly clarifies many of the equilibrium diagrams of alloys of iron, we now give in the case of iron-silicon. Fig. 1 is Stoughton and Greiner's iron-silicon diagram from their American Institute of Mining and Metallurgical Engineers paper.² The present writer's version of this diagram is given in Fig. 2. The Fe-Si is dropped entirely, as not called for by any evidence, and contra-indicated by the X-ray findings. We extend their line KG backwards to B, to put in lines FB and AB.

¹Stoughton and Greiner, Transactions, American Institute of Mining and Metallurgical Engineers, Iron and Steel Division, 1930, p. 160.

²Bragg, Lectures at Cornell University, April, 1934.

Their experiments on electrical resistance versus temperature yield the points along the lines FB and AB indicated by V or Δ .

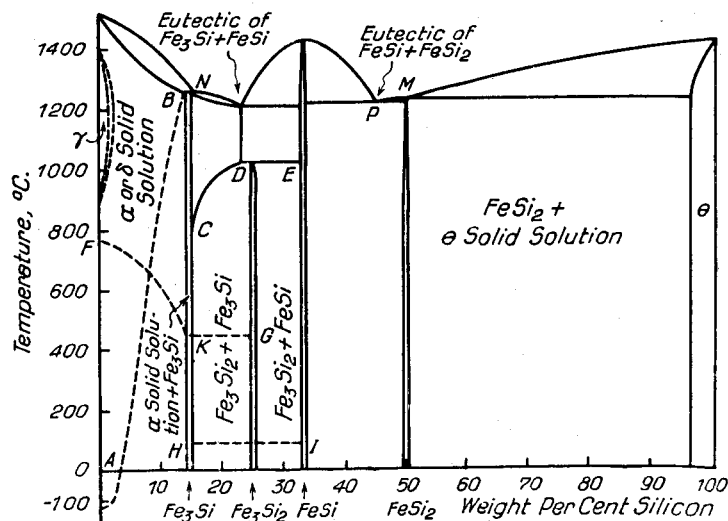


Fig. 1—Stoughton and Greiner Iron-Silicon Diagram (A. I. M. E. 1930).

The V points are from heating curves, and indicate that the equilibrium to which they relate should be at a lower temperature; the Δ are from cooling curves. The author's interpretation is that the point K is a eutectoid point of the β (Si) solid solution; the field GKCDG is two-phased, of β (Si) + Fe_3Si_2 ; the field FBKF is two-phased of α (Si) + β (Si); the field FBA is single-phased, of α (Si).

α (Si), β (Si), and Fe_3Si_2 (?) are all body-centered cubics, very similar in dimensions; this meets the X-ray data. Fe_3Si_2 is formed by a peritectic reaction at D at 1020 degrees Cent. (1870 degrees Fahr.), from β (Si) and FeSi ; but this change is slow and usually incomplete. Hence Fe_3Si_2 is hard to locate by microscope or X-ray, and the magnetic change of β (Si) to α (Si), at 450 degrees Cent. (840 degrees Fahr.), may be traced at times between the composition of Fe_3Si_2 and FeSi . From β (Si) or α (Si), at temperatures 450 degrees Cent. (840 degrees Fahr.) or lower, Fe_3Si_2 precipitates along BKG or AB like Mg_2Si in duralumin, but even quenching is not a fast enough cooling to stop this precipitation, which gives immediate "age hardening" and brittleness. The effect of annealing below ABKG is to soften, and raise the ductility, by aggregating the Fe_3Si_2 . Pilling's data on brittleness of iron-silicon alloys have two causes: the precipitation hardening just

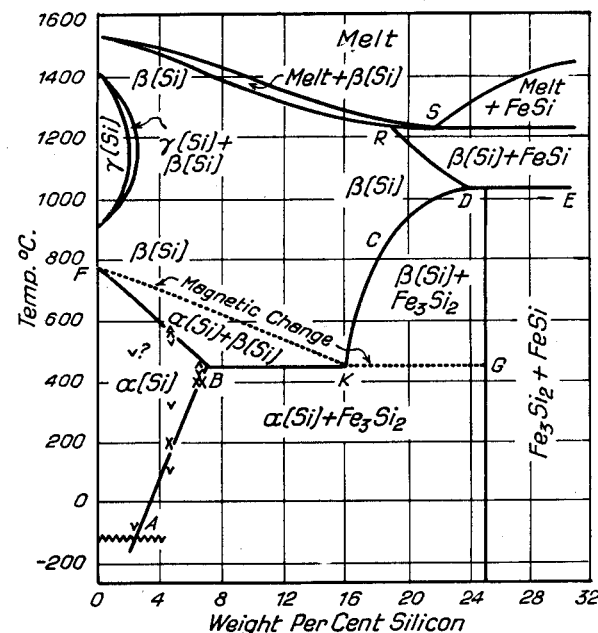


Fig. 2—Upton Diagram of First Part of Iron-Silicon System.

mentioned, below AB, and crossing that, a bit below —100 degrees Cent., and indicated by a jagged line, the general cold-brittleness that comes to most iron alloys at low temperatures. The writer does not believe that the line AB swings sharply to the left just below the point A. Within the α (Si) field, and above —100 degrees Cent., the alloys are ductile.

I submit that as soon as we admit that α (Si) and β (Si) can be and are different phases, we get the iron-silicon diagram in the form of this Fig. 2, clearing up in a consistent whole all the microscopic and X-ray structure, the physical and engineering properties of the silicon steels. Similar diagrams are feasible and useful for all the alloys giving "gamma-loops."

For the description of the behavior of austenite we may discuss first the experiments of T. Sato.* Sato worked with a series of over a dozen nearly pure iron-carbon alloys, covering the carbon range to 1.55 per cent. These steels he heated and cooled through the critical temperatures at various rates of change of temperature, observing the structural changes by magnetometer and dilatometer. He plotted the beginnings (or endings) of A_r and A_c points versus rate of change of temperature, obtaining parabolic looking curves tangent to the axis of zero rate of change of temperature at the equilibrium temperature. These tangencies are

*"On Critical Points of Pure Carbon Steels," Technology Reports of Tohoku Imperial University, 1928.

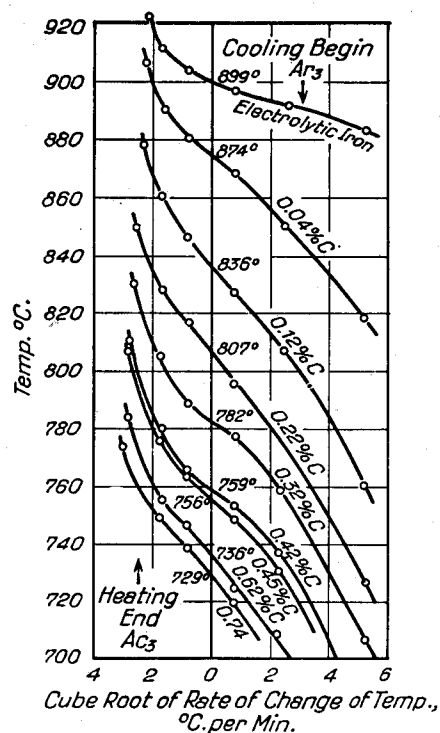


Fig. 3—Sato's Data of Supercooling and Superheating, Hypoeutectoids.

very difficult to estimate with satisfaction. After trials of various relations, the present author found that the superheating or supercooling "lag" of a critical point is quite nearly proportional to the cube root of the rate of change of temperature. In Figs. 3, 4 and 5 are shown Sato's data replotted as temperature versus the cube root of the rate of change of temperature, putting heating data on one side of the axis of zero rate of change of temperature, cooling data on the other side. The equilibrium temperature is then where the common curve from the heating and cooling sides cuts this axis of zero rate of change of temperature, and is now very satisfactorily found. The wildness of some points on these curves is probably due to rate of change of temperature not being quite as reported; the detailed reporting of Sato's data is unsatisfactory in this regard. In Fig. 6 is shown the equilibrium diagram of iron-carbon (for Fe₃C system) as recovered from the Sato data. Note the straightness of the Fe₃C saturation line, the pearlite composition at 0.78 per cent carbon, and the distinctly different lines of separation of beta and alpha iron from austenite.⁴

⁴Compare the substantially equal diagram obtained by another method from Sato's data by Yap, Chu-Pay, Transaction, American Society for Steel Treating, March, 1933. Mr. Yap's method has not the merit of bringing the equilibrium temperature in the middle of the plot, as in Figs. 3, 4, 5 here.

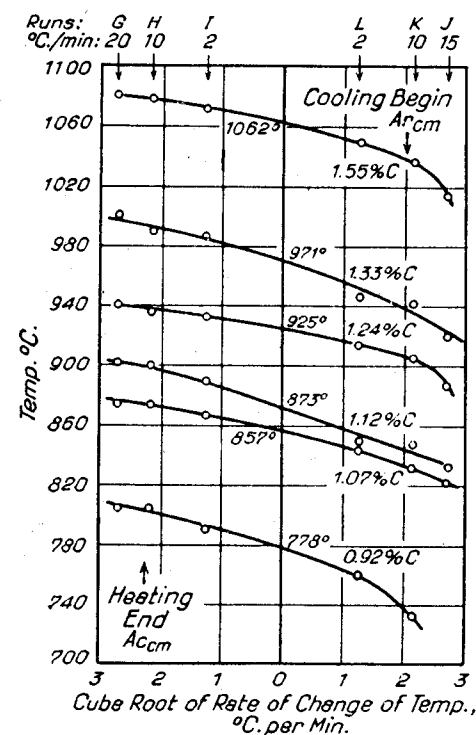


Fig. 4—Sato's Data of Supercooling and Superheating, Hyper-eutectoids.

Sato's cooling rates covered the range from 0.5 degrees Cent. per minute to 150 degrees Cent. per minute. The slopes of the curves in Figs. 3, 4 and 5 give equations for the supercooling "lag" of beginning of precipitation of beta, alpha, Fe₃C, or pearlite from austenite, in the empirical form supercooling = constant times the cube root of the rate of cooling in degrees Cent. per minute. I find no change of these supercooling lag constants with composition of the austenite, except for pearlite below 0.4 per cent carbon. For beta and alpha iron the constant comes out 12+; for Fe₃C 13; for pearlite 9 (except for carbon contents below 0.4 per cent, where the constant increases). It is not surprising that beta and alpha should act alike; but it is rather surprising that Fe₃C should have the same lag as beta or alpha. Most remarkable is the lesser lag for pearlite as compared to its components alpha and Fe₃C individually. The consequences of this are shown in Fig. 7. The upper solid lines of this Fig. are the equilibrium diagram lines of the bottom of the austenite field. The parallel sets beneath are the lines of beginning of precipitation from supercooled austenite of beta, alpha, P (pearlite), or Fe₃C, for a number of successively increasing rates of cooling. Note that as a consequence of the lesser supercooling of pearlite than of its components the composi-

tion of pearlite is a point only at equilibrium; at any finite rate of cooling, that is, in any practical case, pearlite has a range of com-

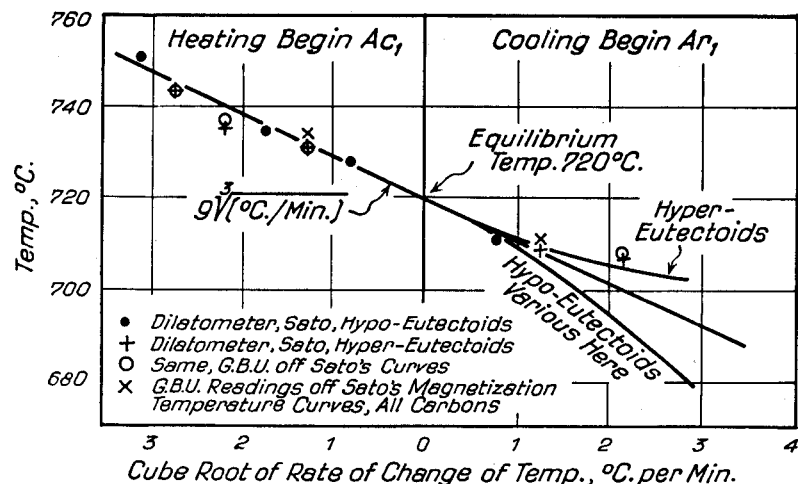


Fig. 5—Sato's Data; Critical Temperature, Superheating and Supercooling of Pearlite.

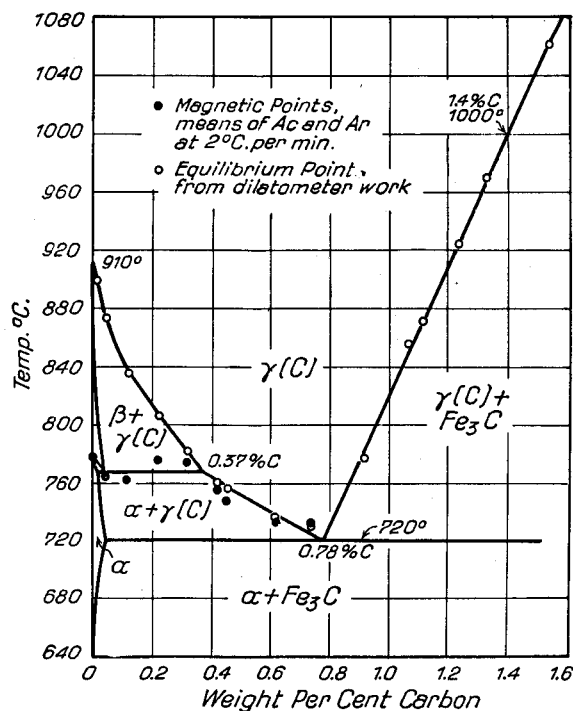


Fig. 6—Fe-Fe₃C Equilibrium, Drawn by Upton from Sato's Data.

positions from where its lesser supercooling eats up the end of the alpha line on one side to where it eats the end of the Fe₃C line on the other side. The cooling rates shown in Fig. 7 go far beyond Sato's range, and the extrapolation may not be safe numerically, but in principle it should be safe. The faster the cooling rate, the wider is the range of composition of pearlite, particularly on the low carbon side of the equilibrium eutectoid. We are all familiar in practice with the increase of pearlite area in a cross section and decrease of ferrite, in a medium carbon steel cooled at increasing rates; but the author believes that this is the first theoretical explanation of that familiar phenomenon. Note that at the cooling speeds of water quenching pearlite would extend well over toward pure iron; or in other words, the difference in equilibrium temperatures between iron and pearlite precipitations from austenite would be wiped out by the difference between supercooling lags. Again, we have known that on cooling curves Ar_{3.3} and Ar₁ points merged as cooling rates increased, but have not usually recognized that such merging resulted in such a showing as Fig. 7. The final consequence is that at high cooling rates, as in heat treatment quenching, we need look only for the pearlite precipitation out of the austenite, for the alpha and Fe₃C

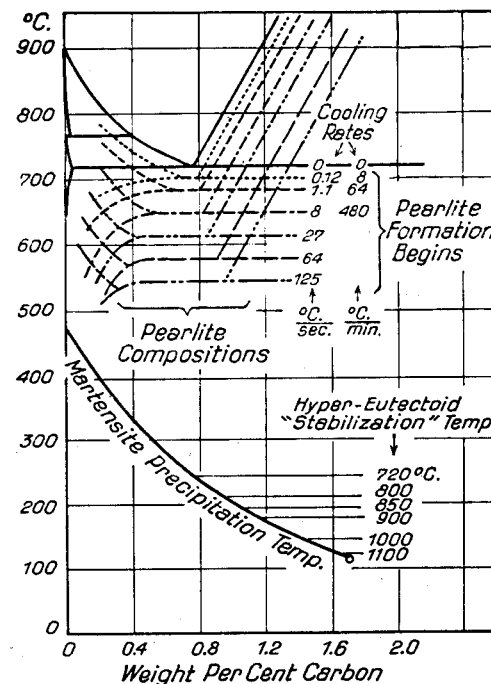


Fig. 7—Composition and Precipitation Temperatures of Decomposition Products of Austenite (Pearlite and Martensite) at Various Cooling Rates, and After Various Stabilizations.

precipitations will be merged in that; and such pearlite will have the composition of the austenite, not eutectoid composition.

While we are studying Fig. 7, we may see in its lower part the critical temperatures at which martensite precipitation begins during cooling in carbon steels. These temperatures are of a reaction in supercooled austenite, and the temperature depends on the amount and nature of the alloy in solid solution in the austenite; in this case practically carbon only. Inspection of the curves of Davenport and Bain⁶ shows that at 550 degrees Cent. (1020 degrees Fahr.) carbon steels require 0.5 to 1.0 seconds to start pearlite formation, and 5 to 10 seconds to complete it. Below 500 degrees Cent. (930 degrees Fahr.) these required times increase with great rapidity as the temperature falls. Hence it is entirely safe to say that, during a continuous cooling of a steel, if the

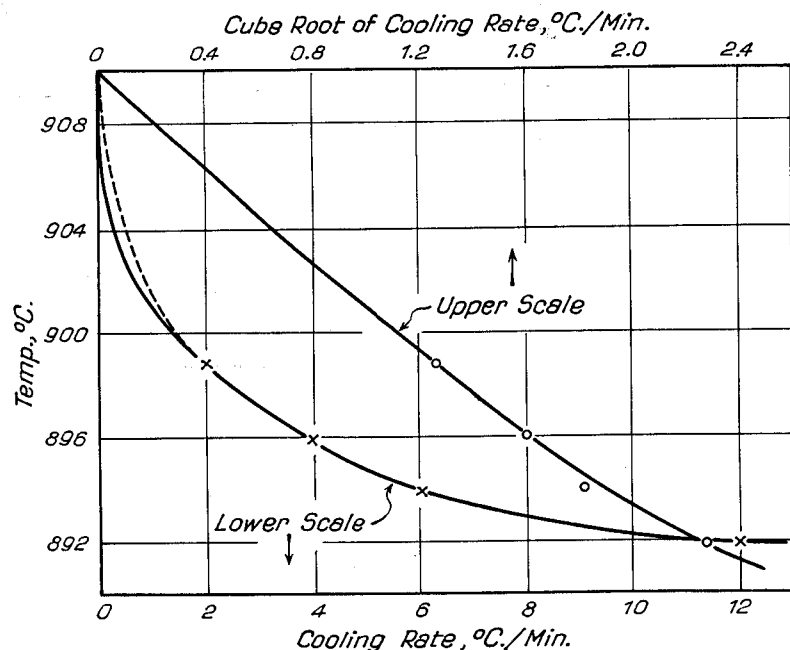


Fig. 8.—Data of Ruer and Klesper on Supercooling in "Pure" Iron.

pearlite change has not occurred before the steel passes 500 degrees Cent., that change is suppressed and the austenite supercools to and below the martensite critical temperatures. The martensite formation becomes rapid within 50 degrees Cent. drop below the martensite critical temperatures. It does not follow by logic, but it seems plausible to expect, by downward extension of the way the composition of pearlite spreads out as its temperature of formation lowers, that martensite would have the com-

position of its preceding austenite. This is now generally admitted to be the case.

In Fig. 8 we have applied the empirical law of supercooling varying with cube root of cooling rate to data of Ruer and Klesper⁷ taken from National Metals Handbook. The original authors, plotting the data versus first power of the cooling rate, getting a parabolic curve tangent to the temperature axis, guessed 906 degrees Cent. (1663 degrees Fahr.) as the equilibrium transition temperature of $\gamma \rightleftharpoons \beta$ change. Fig. 8 shows it should have been about 910 degrees Cent. (1670 degrees Fahr.). The slope in Fig. 8 is 8+ when cooling rate is in degrees Cent. per minute; Sato's steels gave 12+, except that for electrolytic iron he had 3+. Here we find the *form* of the empirical law is general, but the *constant* is variable as we change the source or preparation of the sample.

The variation of the supercooling lag constant with source and preparation of sample is one aspect of the familiar phenomenon known in heat treatment practice as "stabilization" of the austenite. "Stabilized" austenite is slower to react in cooling than unstabilized; in the present language of this paper, it has a greater constant of supercooling lag. There are three actions concerned in stabilization. Due to superheating lag, analogous to supercooling lag, we have to heat a steel considerably over the equilibrium lower boundaries of the austenite field before it becomes entirely austenitic; i.e., loses all vestiges of the low temperature forms α , P. Fe₃C. But when γ (c) is first fully formed it is not internally homogeneous as to composition. To make it homogeneous requires time for diffusion of iron and/or carbon back and forth—half an hour to even several hours of holding at constant temperature. If we do not give diffusion time the parts low or high in carbon in the austenite form nuclei of α or Fe₃C earlier on cooling than would be the case if we had given time for diffusion to equilibrium—the unhomogeneous austenite reacts faster than homogeneous austenite. Quite inseparable in practice from the effect of departure from homogeneity is the factor of grain size (or crystal size) of the austenite. The boundaries of a crystal are usually and inherently more active than its core when changes are starting (except when non-homogeneity is great); hence the finer the austenite grain size the faster does it react on cooling. Reaction rate is perhaps proportional to number of grains per cubic inch.⁸ Lastly, and particularly in tool steels and alloy steels (whenever we are dealing with the right hand boundary of the austenite field, the saturation lines for carbon or for other alloying elements) we can change the composition of the solid solution austenite by choice of the holding temperature for stabilizing the austenite. By increasing the content of the austenite solid solution as to carbon, vanadium, manganese, nickel, chromium, tungsten, etc., we can marked-

⁶Ferrum, 1914.

⁷Deduced from data of Bain, Campbell Memorial Lecture, Transactions, American Society for Steel Treating, 1932.

⁸Transactions, American Institute of Mining and Metallurgical Engineers, Iron and Steel Division, 1930.

ly slow up the reaction rate of that austenite when we cool it; and in many cases we also change the critical temperatures at which such stabilized high-alloy austenite reacts in cooling—notably the martensite point which depends mainly, as the pearlite point does not, on the composition of the austenite. See the bottom part of Fig. 7 for this effect of stabilizing temperature on martensite critical temperature, and for carbon as alloying element.

We now follow out the consequences of the supercooling lag laws into wider fields. The iron-carbon system has two alternative equilibria, the Fe-Fe₃C and Fe-G systems. The Fe-G system is the more stable, the slower to form; it is a generality of chemical reactions that the metastables are faster than the stable reactions where both are possible. A. E. White, some years ago,⁸ stated that when malleable iron is being made the pearlite eutectoid occurs on cooling if the cooling rate exceeds 5 degrees Cent. per hour; the Fe-G eutectoid occurs only when the cooling rate is less than 4 degrees Cent. per hour. The temperature separation of the two eutectoids is here perhaps 10 degrees Cent., the Fe-G at the higher temperature. Obviously, the supercooling lags of the two eutectoids are so related to each other that up to cool-

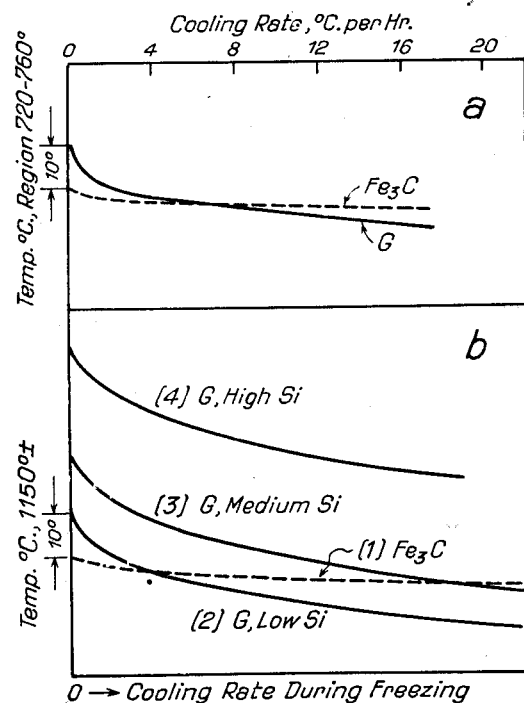


Fig. 9—Relations of Stable Fe-G and Unstable Fe-Fe₃C Systems at Eutectoids (a) and Eutectics (b).

⁸Transactions, American Society for Steel Treating, 1926, p. 214.

ing rates of 4 degrees Cent. per hour the G eutectoid occurs first; above 5 degrees Cent. per hour the Fe₃C eutectoid occurs first. With Sato's supercooling lag constant for pearlite the supercooling of that eutectoid at 5 degrees Cent. per hour cooling rate is about 4 degrees Cent. At this same cooling rate the stable Fe-G eutectoid must be supercooled $10 + 4 = 14$ degrees Cent.; hence its supercooling lag constant must be of the order of 3.5 times as great as that for pearlite. The relations of the two eutectoids are shown in Fig. 9a. As soon as the cooling rate is great enough to bring the supercooled G eutectoid temperature close to the (less) supercooled Fe₃C eutectoid, the more rapid action of the latter will dominate. Since we almost never cool steels, even in furnace cooling, at rates slow enough (less than 2 degrees Cent. per hour) to give the G eutectoid a chance, we practically never see any but the pearlite eutectoid.

The probable scheme of action of cast irons at the eutectic (freezing) temperatures, and possible scheme at the eutectoid temperatures, is seen in Fig. 9b. Silicon raises the equilibrium temperature of the G eutectic considerably (15 degrees Cent. for each weight per cent Si) and has practically no effect on the temperature of the Fe₃C eutectic. The separation of the two eutectics, at equilibrium, is not over 10 degrees Cent. at low silicons, but increases relatively sharply as silicon rises. Curve (1) in Fig. 9b represents the temperature of beginning of Fe₃C eutectic versus cooling rate for all silicons; curve (2) the corresponding G eutectic at low silicon; (3) the G eutectic at medium silicon; (4) the G eutectic at high silicon. Obviously, at low silicon we will ordinarily get the Fe₃C eutectic, at medium silicon we will ordinarily get the G eutectic, but by chilling (very rapid cooling) we can get the Fe₃C eutectic and with high silicon we cannot make white cast iron even with fairly drastic chill. Added to relations shown in Fig. 9b one other factor nearly completes the statement of basic relations of gray and white cast irons to each other at freezing. This other factor is the latent heat of freezing, which rises with either carbon or silicon increase. When the latent heat is high it becomes difficult to force rapid cooling during the eutectic freezing. The laws of supercooling of solutions then explain the relations of white and gray cast irons. "Mottled" cast irons are of course those where the cooling rate is such that the two eutectics are both forming simultaneously during the freezing, at comparable rates.

The statement of an approximate empirical law of supercooling, and showing of its power to explain much in heat treatment relations in steels and cast irons, is not a stopping point of our study if we can make more detailed analysis of what lies behind the empirical approximation. This deeper analysis is now becoming possible. Some of the Germans are already attempting analyses of the laws of chemical reactions in the solid. We have in this country, in the Davenport and Bain paper on "Transformation of

Austenite at Subcritical Temperatures," when we back that up with other papers in world literature of metallography of the last ten years or so, a fairly adequate basis for beginning of a quantitative analysis of heat treatment reactions, which are simply chemical reactions in solid crystalline material, occurring over a far wider range of temperatures than a chemist usually has had to deal with. The law of supercooling, with which we have been dealing, is but one small part of this wider field now opening up before us.

Davenport and Bain, in the paper just mentioned, worked with five kinds of carbon steel, one low alloy and one high alloy chromium steel. Numerous samples of each steel were heated in the lower part of their austenite fields, as for heat treatment hardening, and then suddenly transferred to quenching baths thermostatically controlled. The bath temperatures covered in many steps the region from room temperature to the pearlite critical,

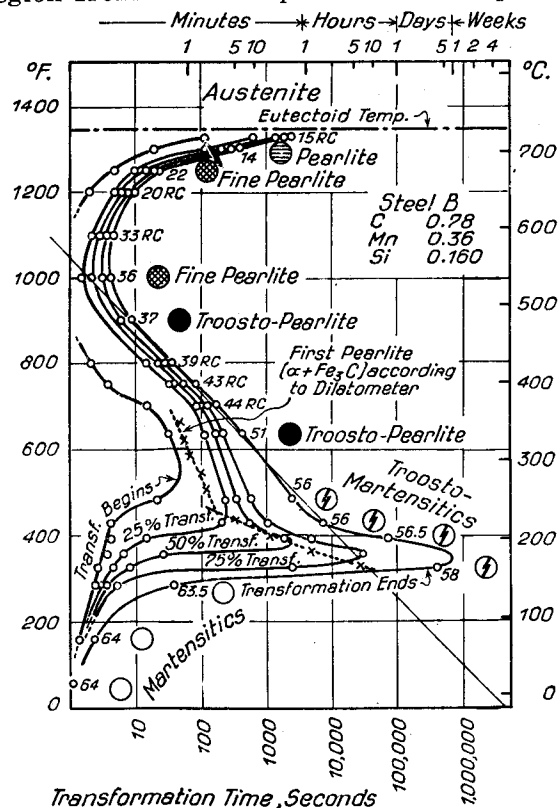


Fig. 10—Decomposition of Austenite at Constant Temperature, at Various Temperature Levels under the A_1 Critical Temperature (Davenport and Bain With Additions by Author).

⁹Transactions, American Institute of Mining and Metallurgical Engineers, Iron and Steel Division, 1930.

A_1 , of each steel. After various lengths of stay in the quenching bath, from seconds to months, specimens were withdrawn, water-quenched, and examined microscopically for amount and kind of change undergone. At some lower temperatures a dilatometer was clamped onto the specimen as it went into the quenching bath and kept on to record dimensional changes versus time. Results were plotted with temperatures of quenching baths as ordinates, with contour lines of logarithms of time to certain decimal amounts of change of austenite toward its final steady condition at the bath temperature used. One of their figures is reproduced here as Fig. 10.

A survey of all of their figures for carbon contents 0.5 to 1.2, confirms the spread of composition of pearlite with increasing supercooling as shown in Fig. 7, but with less spread toward alpha and more toward Fe_3C than there indicated. They noticed little or no proeutectoid carbide even with carbon at 1.13 and 1.17 per cent. The cause of this most probably is that proeutectoid carbide must be large in amount to be observed as such, that small amounts may be lost in pearlite.

One feature of Davenport and Bain's charts of log-time for various percentages of transformation versus temperature catches

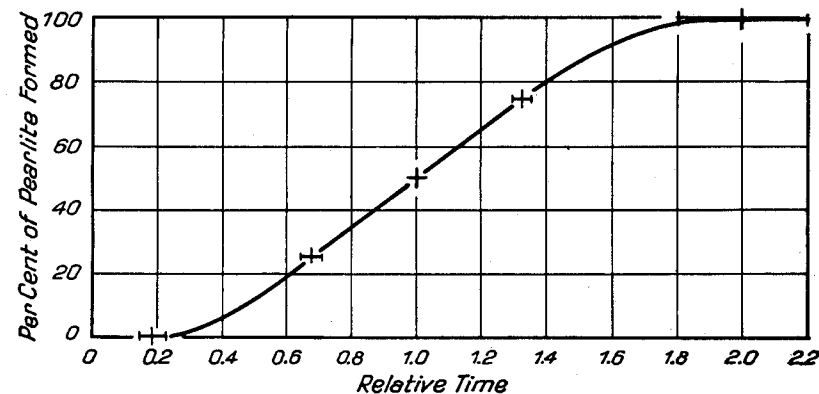


Fig. 11—Form of Time History of Pearlite Precipitation from Supercooled Austenite, at All Temperatures Where it Occurs.

the eye immediately. It is that the contour lines for equal percentages of formation of pearlite are parallel over the temperature range from the pearlite critical temperature to the martensite critical temperature. That is to say, the relative internal history of pearlite formation is the same over all this temperature range; the only change is in the time scale of time required. Hence it is possible and permissible to average the log-time versus per cent change data for all temperatures in the range mentioned to find for each steel the typical form of the pearlite transformation versus time. The numerical results of this averaging are shown in Table I, and plotted in Fig. 11.

Table I

Steel	Composition, Weight Per Cent				Critical Temperature °C				Relative Reaction Time for Percentage of Formation of Pearlite				
	C	Mn	Si	Cr	P	M	P	M	0.5	25	50	75	99.5
A	1.13	0.30	0.17	0	728	190	1345	375	0.19	0.71	1.00	1.32	2.00
B	0.78	0.36	0.16	0	728	250	1345	480	0.15	0.67	1.00	1.32	1.85
C	0.54	0.46	0.20	0	728	300	1345	570	0.21	0.70	1.00	1.39	2.24
D	0.50	0.91	0.13	0	724	285	1335	550	0.32	0.74	1.00	1.53	2.56
E	0.64	1.13	0.09	0	724	260	1335	505	0.16	0.65	1.00	1.31	2.30
F	1.17	0.30	0.18	0.26	735	170	1355	340	0.28	0.74	1.00	1.31	2.04
G	0.38	---	---	13.	815	240	1495	470	0.13	0.63	1.00	1.25	1.53
Average without D									0.19	0.68	1.00	1.32	1.99

The relative times have been averaged for all the steels except D, because that one seems to have mislaid its 50 per cent time relative to its other times. Fig. 11 plots these average results as the type of relative reaction time at constant temperature for various structural percentages of pearlite to be formed from austenite. The effect of different temperatures, used as the temperature locus for this pearlite formation, is simply to change the actual amount of time corresponding to the unit of time in Fig. 11. Unit time is taken at the 50 per cent change point because in that region the times can be much more accurately fixed than at the 0.5 and 99.5 per cent points, where the curve is too flat for accurate time measurement. The mean time deviations from the averages are indicated in Fig. 11, the unity at 50 per cent change having been forced.

It will be noted in Fig. 10 that this formation of pearlite is all that the steels do between the pearlite critical temperature and the martensite critical temperature. The only effect of temperature change in that interval is to change the time required for the pearlite formation. As temperature falls this time comes down from infinity at A_c to a few seconds at around 1000 degrees Fahr. (540 degrees Cent.); then lengthens rapidly as temperature falls further. Nevertheless, we have not been calling all the results of this pearlite reaction pearlite. Below the temperature of minimum time of reaction, or maximum rate of reaction, the product has commonly been called "nodular troostite." It grows more or less spherically from a few nuclei which are commonly found along crystal boundaries of austenite. Inside each nodule we now know, from the work of Lucas and of Davenport and Bain, the structure is that of pearlite, simply much smaller in crystal size than the pearlite formed at higher temperatures. There are two other structures in steels also called troostite. One of these is a ternary eutectoid of alpha iron, iron carbide, and alloy carbide, forming at about 400 degrees Cent. (750 degrees Fahr.) in such low alloy steels as 3.5 per cent nickel, 1.7 per cent chromium.¹⁰ This ternary eutectoid may form in addition to a regular pearlite in such steels; because of the low temperature of formation it is nodular. The third troostite is the first product of decomposition of martensite in tempering. The author ventures to suggest that

¹⁰Andrew, Rippon, Miller and Wragg, Journal, Iron and Steel Institute, 1920, I.

the first two kinds of "troostite" ought to be called "troostopearlite," to make clear at once their appearance and their real nature; while the last one only might properly be called troostite. It might also be added that the presence of Fe₃C in the real troostite is quite dubious. Its specific volume and electric resistance are both higher than for sorbite, into which troostite goes by further tempering; and both excess specific volume and excess electric resistance of troostite over sorbite, plotted against carbon content, give straight lines joining with the sorbite properties at zero carbon, showing that alpha iron is present in both alike, but the carbon condition is different. Further, magnetism (flux density in strong field) decreases at over 200 degrees Cent. temperature when troostite is tempered to sorbite, corresponding to using up of hitherto alpha iron in forming of carbide, which at those temperatures is non-magnetic.¹¹

Now consider the variation of rate of formation of pearlite with temperature between the pearlite and martensite critical temperatures. There are two temperature factors involved. One is the effect of absolute temperature on all chemical reactions, making them slower as temperature falls. The other varies with the temperature drop below the critical temperature, and tends to make the reaction faster as the temperature drop increases. The interaction of the two temperature effects gives the minimum of reaction time noted for pearlite formation at around 1000 degrees Fahr. (540 degrees Cent.). We will try to untangle the two effects, which we may call ϕ (T) and ψ (T-T_c). The ψ (T-T_c) may be peculiar to this reaction, though its type is general; the ϕ (T) is a general function.

In text books of physical chemistry one finds that time rate of reaction, over wide ranges of temperatures, is given by the form

$$\log R = -\frac{A}{T} + B + CT,$$

where R is reaction rate, A, B, C are constants. For lower temperatures the CT term is commonly omitted; but for high temperatures it becomes, in the end, the dominant term of the expression. Searching for some convenient way to show how this equation works out, the writer has picked, as perhaps the easiest to visualize, the plotting against absolute temperature of the size of the increment of temperature required to double reaction rate. Differentiating the expression for reaction rate with regard to temperature, and remembering that $\log 2 = 0.693$, one gets the increment of T to double reaction rate = $\Delta T = 0.693^2 / (A + CT^2)$. At low temperatures this gives ΔT varying as T^2 , for CT^2 is negligible; but at high temperatures the expression approaches $0.693/C$, a constant. Hence the form of the relation of ΔT to T is as shown in Fig. 12. On this Fig. we spot a few things we know, as that at room temperature chemical reactions commonly double in rate

¹¹Matsumita and Nagasawa, Journal, Iron and Steel Institute, 1927, II.

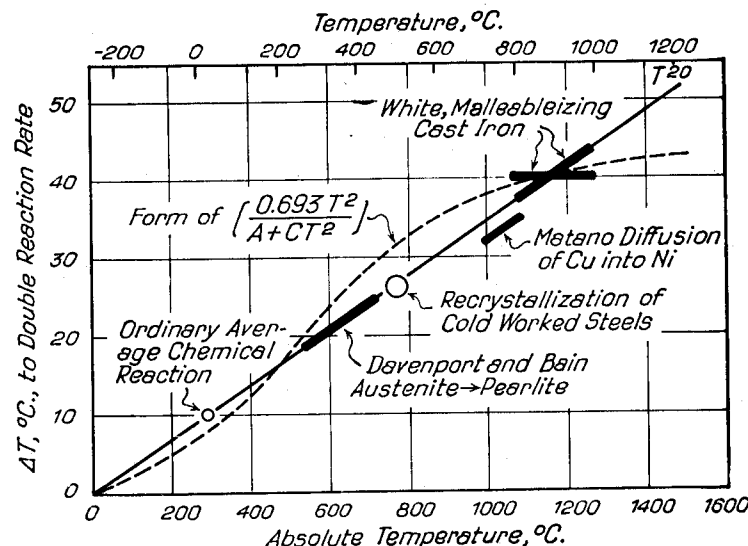


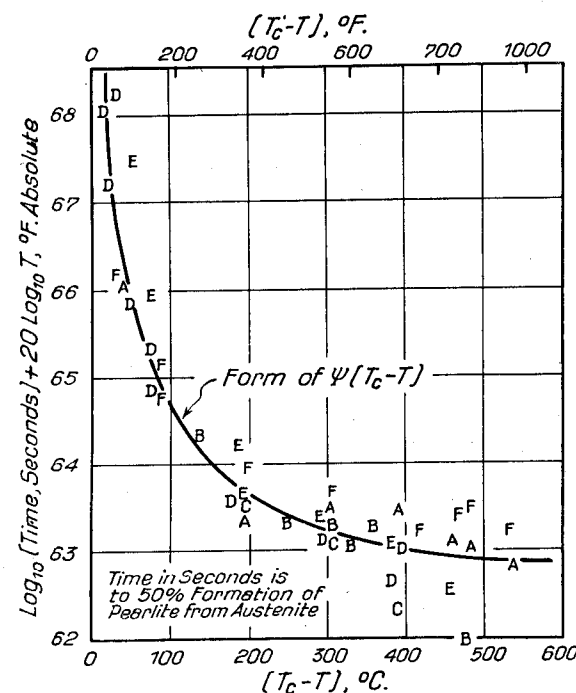
Fig. 12—General Effect of Temperature on Reaction Rates.

for about a ΔT of 10 degrees Cent.; at about 500 degrees Cent. ΔT to double rate of recrystallization of cold-worked steel is 25 degrees Cent.; in malleableizing cast iron ΔT is 40 degrees Cent., at temperatures 1500 to 1900 degrees Fahr. (820 to 1040 degrees Cent.).

We need to be careful as to what and why we choose for values of this ΔT . In general, reactions in the solid require atoms to move from here to there, to get together with their new partners and build the new crystals. Hence, basically, reaction velocities in the solid are conditioned by diffusion velocities. ΔT is controlled not by the absolute value of diffusion velocities but by the effect of temperature on relative increase of those velocities. Take for example the recent work of R. Schneidewind and A. E. White on malleableizing.¹² As soon as Fe_3C begins to break down, it forms γ (c) between the G and the Fe_3C , and the solid solution thenceforth transfers carbon atoms from decomposing Fe_3C to the G particle which is growing. The driving force is the greater solubility of carbon from Fe_3C than from G in the γ (c); the speed of diffusion depends on the distances between nuclei. Plots of reaction time against temperature in the Bulletin give 40 degrees Cent. = ΔT to double reaction rate; but they fit equally well the twentieth power of absolute temperature when reaction time is plotted logarithmically versus absolute temperature. A survey of Fig. 12 raises question whether ΔT should be represented by an S shaped curve, or by a straight line through the origin. If $R = mT^n$, then $\Delta T = 0.693T/n$, a straight line through the origin with slope $0.693/n$. Since it seems likely that pearlite formation rate de-

pends on diffusion through austenite just as malleableizing does, we have plotted Davenport and Bain's curves of log-time versus temperature in the form of log-time versus log-temperature; and find that beginning a bit below the temperature of minimum time, and down to the temperature of martensite formation, the slope of the log-log plot is about 20 for their steels A to F inclusive; G is not available. Hence we feel it justified for the present at least to use the approximation $\phi(T) = mT^{20}$, for it appears valid from 300 to 550 degrees Cent., and again from 800 to 1050 degrees Cent., and can hardly fail to be valid between those two ranges and also fairly well outside of them.

Multiplying the times of reaction for various steels for pearlite formation at various temperatures by T^{+20} leaves two parts, a constant specific to the reaction, and the function $\psi(T_c - T)$. The results are shown for the different steels in Fig. 13, as $\log_{10} [\text{constant} \times \psi(T_c - T)]$ versus $(T_c - T)$. The hyperbolic form of these curves suggests strongly that $\log \psi(T_c - T)$ approaches zero or a very small value as $(T_c - T)$ increases. There is no point in trying to push this analysis of $\phi(T)$ and $\psi(T_c - T)$ much further, for each depends on the other here, and independent accurate knowledge of $\phi(T)$ is needed before $\psi(T_c - T)$ can be really evaluated.

Fig. 13—Form of the Function $\psi(T_c - T)$ the Effect of Distance Below its Equilibrium Critical Temperature on Time Rate of a Reaction.

¹²University of Michigan Engineering Research Bulletin, August, 1933.

The general form of the $\phi(T)$ and $\psi(T_c - T)$ functions is well enough shown, however, to form the basis of an important argument. It seems incredible, in view of the forms of these functions, that $\psi(T_c - T)$ should suddenly throw a fit, so to speak, at the martensite critical temperature, so that martensite could be explained as a sudden, vastly more rapid, precipitation of $\alpha + \text{Fe}_3\text{C}$ (troostite-pearlite). Yet such an explanation of the A_r critical temperature of steels has often been offered. Using the $\phi(T) = T^{20}$ to approximate the effect of absolute temperature, to make comparison of the minimum times of formation, of martensite with those of pearlite, as of the same temperature, it comes out from the Davenport and Bain data that martensite formation has a reaction rate from a billion to a trillion times as fast as pearlite; the jump required in $\log_{10} \psi(T_c - T)$ is 9 to 12.

The X-ray shows martensite to contain a tetragonal form of iron, not the cubic alpha; the electrical resistance is as high as that of gamma iron, not low like alpha; the specific volume is 3 per cent or more greater than alpha, agreeing with the X-ray data for the tetragonal iron; magnetism increases when martensite tempers to troostite, corresponding to the change of tetragonal iron to cubic alpha, and indicating that tetragonal iron (real martensite) is non-magnetic. Iron carbide can be looked for by its characteristic reversible loss of magnetism between 150 and 200 degrees Cent.; martensite and troostite do not show this, but sorbite does. The implication here that Fe_3C is not present in martensite and troostite cannot be dodged by argument about submicroscopic size of the carbide; all the evidence about relation of size of carbide to magnetic properties shows it more strongly magnetic as its crystal size decreases; as if, perhaps, the carbide had adsorbed alpha iron on its surfaces. Do not assume that a statement that tetragonal iron (real martensite) is probably non-magnetic means that martensitic steels are nonmagnetic; the maximum amount of tetragonal iron is probably never over 35 per cent of the whole, and most of the rest is alpha except for carbon contents toward 1.5 and fairly drastic quenches from very high temperatures.

Other very convincing evidence that martensite formation cannot be a suddenly revived, enormously speeded pearlite reaction is found in the steels of nickel 3.5 per cent, chromium 1.7 per cent.¹³ These can be stabilized to show, in the same specimen in one single cooling, pearlite at 600 degrees Cent., ternary eutectoid at 400 degrees Cent., and martensite at 150 degrees Cent. Since in one specimen in one cooling pearlite cannot be both supercooled to 150 degrees Cent. and not supercooled below 600 degrees Cent., this experimental showing entirely wrecks the theory that martensite is the same as pearlite, but forming rapidly at a low temperature when stopped at a high one. Yet martensite must be an iron phase, for the critical temperatures of martensite formation

¹³Andrew, Rippon, Miller, and Wragg, Journal, Iron and Steel Institute, 1920, I.

shown in Fig. 7 slope down with increasing carbon just as do the beta and alpha lines below the γ (c) field above. Martensite has its own $\psi(T_c - T)$, its own critical temperatures, independent of alpha iron. That martensite is tetragonal iron, an unstable allotrope, if you wish, different from alpha iron, seems an inescapable conclusion. Whether the carbon in martensite is in solid solution can be answered only by the final settlement, now in doubt, whether the space lattice shown by the X-ray changes with increase of carbon content.

We return now to another Fig. of the Davenport and Bain paper, reproduced here as Fig. 14, showing hardness of completed transformation products for the different quenching bath temperatures. Note that there is no jump in hardness when martensite comes in; simply a progressive increase of hardness in going from pearlite to fine pearlite to troostite-pearlite to troostite-martensitic to martensitic. In the range above the martensite critical temperatures there is no question that increasing hardness is due to increasing smallness of average crystal size. Is it not then reasonable, in view of the continuity of these hardness curves against temperature of structural change, to say that in the martensitic structures also the hardness is due to still finer average crystal size? In this connection we have plotted elsewhere data of hardness of alpha brass versus number of crystals per linear inch, a presentation which plots substantially a straight line; extrapolation of that line to the number of crystals per inch indicated by X-rays for martensite would predict brass of that crystal fine-

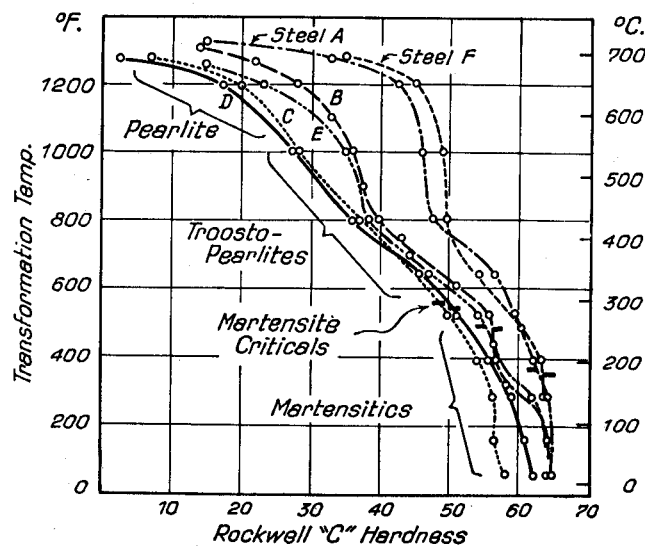


Fig. 14—Davenport and Bain's Figure of Hardness of Completed Transformation Products of Constant Temperature Transformations of Austenite, with the Author's Marking of Martensite Critical Temperatures of Each Steel Added to Their Figure.

ness to have Brinell 3000! There is no need to call upon any factor but crystal size to account for the hardness of hardened steels.

In Fig. 15 are reproduced the dilatometer curves of Davenport and Bain for their steel B, the same steel to which Fig. 10 applies. Note in the upper part of Fig. 10 the pearlite formation, of form as given in Fig. 11. Then, in the lower part of Fig. 15, note the very different habit of formation of martensite. Pearlite has a considerable time lag before any pearlite shows; martensite forms immediately and most rapidly at the start. In form of transformation amount versus time, martensite has the form which a physical chemist would call the mass law form for a first order reaction. If A represents austenite, M martensite amounts, the shape of the martensite transformation curve corresponds mathematically to such a scheme $dA/dt = -RA$ at constant temperature; $dM = -dA$. There is no time lag for martensite as there is for pearlite; consequently, as experiment indicates, martensite formation cannot be supercooled.

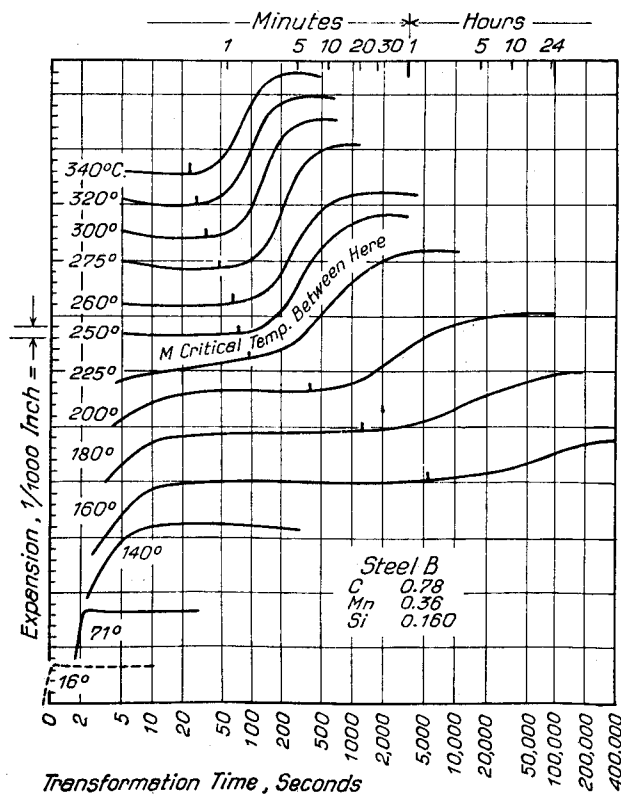


Fig. 15—Dilatometer Curves of Transformation of Austenite at Constant Temperatures. Davenport and Bain Figure with Additions by Author as to Martensite Critical Temperature and Beginning of Pearlite.

Pearlite formation has that time lag at the start which makes supercooling possible. All solution decompositions which can be supercooled, whether of liquid or solid solutions, must be of the pearlite type.* They appear to belong to what the chemist calls sequential or chain reactions. In this case, austenite changes to X_1 , this in turn to X_2 , this in turn to pearlite; I do not intend here to say just how many intermediate, unknown or unrecognized stages there must be between austenite and pearlite, but there must be one and may be two or three. Sato's experiments with dilatometer and magnetometer make it clear that up to the first appearance of pearlite in supercooled austenite the austenite does not reveal what is going on within it, as X stages, which is to produce pearlite as an end product.

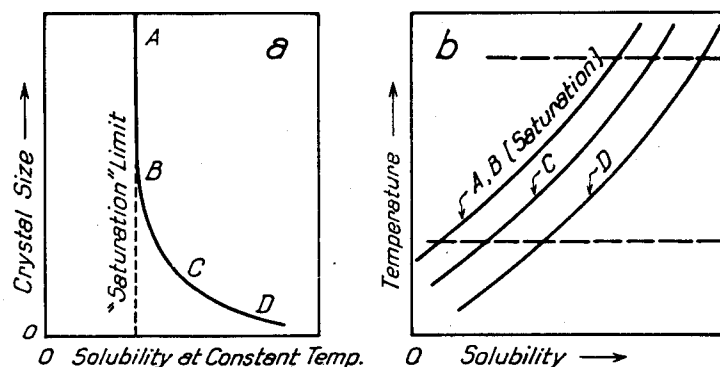


Fig. 16—Solubility Relationships as Affected by Crystal Size.

One part of the X stages in the chain which ends in pearlite is the formation of nuclei in the austenite, of things about to be from decomposition of the austenite. This nucleus formation tangles up in the variation of solubility which is a function of crystal size, as shown in Fig. 16a, and 16b. In Fig. 16a solubility at constant temperature is plotted against crystal size, in the curve ABCD. For large crystals, at A or B, solubility is fairly fixed, and this solubility of large crystals is what we know commonly as the "saturation" limit for the solution. But when crystals are made smaller they are found to be more soluble as their size decreases, as at C and D. This determination is a common enough experiment in chemistry, but we do not know where the curve goes to when the crystal size becomes submicroscopic, below D. The principle involved comes into the growth of crystals; if both large and small ones are in contact with a common solvent, the small ones will dissolve and the dissolved material precipitates out on the large ones, making them grow at the expense of the small ones. We are familiar with such growth of crystals in metals reheated

*The textual argument following is correct if reactions in the solid follow the kinetics of reactions in gases or liquids. But conceivably reactions in the solid require (a) a time rate of formation of nuclei, (b) a time rate of growth upon the nuclei, complicated by mutual interferences of such growths. This would make a story of reaction mathematically quite different from that in gas. See discussion by Dr. Mehl.

after working. Another aspect of this variation of solubility with crystal size is shown in Fig. 16b. The ordinary saturation line is AB, for larger crystals. Smaller crystals have each a solubility line appropriate to their size, like C and D. The letters in Fig. 16a and Fig. 16b correspond to each other, A in one to A in the other, etc.

Now apply the solubility-crystal size relationship to nucleus formation in a supercooled solution, which means one cooled below its saturation line. We may view austenite as a solution of alpha iron and carbon in gamma iron. The kinetic heat motions of atoms in the space lattice may result in change of electron structure of gamma atoms to make temporary alpha atoms; other chance motions may release carbon atoms from solution. But such release cannot last at small supercoolings, for the released material is smaller than D while the solubility conditions are in the region near AB, Fig. 16b; the too small stuff released will redissolve. Not only must atoms of alpha and/or carbon be released, but big enough groups of such atoms be formed and released so their size will be greater than the solubility limit size corresponding to the temperature at which they are released. Hence at temperatures below but near the saturation line nuclei are few, large, and slow to form; at a lower temperature nuclei are more numerous, smaller, quicker to form. This accounts for the $\psi(T_c - T)$ function which quickens precipitations from the supercooled solutions as the temperature level drops. Then comes in the general effect of absolute temperature in slowing up heat motions of heat atoms, and slowing diffusion; ultimately perhaps decreasing the number of nuclei forming.

After nuclei of alpha iron and carbon have been formed as a start toward pearlite, reaction must take place between the carbon and some form of iron to make Fe₃C. The formation of Fe₃C is faster than the growth of graphite. The formation of alpha iron from gamma requires considerable movement of iron atoms from their positions in the gamma space lattice. In nucleus formation, carbide formation, and growth of crystals of alpha and Fe₃C, we may have found enough X stages to make up the chain of reactions in pearlite formation. The more intermediate steps there are in such a chain, the later in the total reaction time do the first signs of the final product appear; and the time history of pearlite formation, as shown in Fig. 11, has so late a start of pearlite showing as to indicate at least two intermediate stages.

Martensite (tetragonal iron) formation from gamma iron requires very little atom motion to take up the new configuration. Hence it can be a fast action even at low temperatures. The octahedral planes of densest packing of iron atoms in austenite are also natural planes of similar packing in the tetragonal iron, so that growth of martensite along such planes is very easy and quick. Hence the structure of martensites. And hence also its refusal to supercool; its lack of delay stages of chain reactions.

Martensite has a trick of action which is peculiar to reactions in the solid involving considerable volume changes. At low temperatures austenite is 2.5 to 3 per cent more dense than the pearlite condition of the same steel, but martensite is 3 per cent less dense—a density or volume jump from austenite to martensite of 5.5 to 6 per cent, an expansion when martensite forms. In a liquid or a gas pressures must be hydrostatically equal throughout the mass during a chemical reaction, except with detonations; but in a rigid solid great local, point to point almost, variations of pressure are possible. A compression stress will stop martensite formation from austenite, stabilize the austenite. Hence martensite formation, as is well known, behaves differently during cooling of a piece of metal and during holding at a steady low temperature. At steady temperature the martensite formation is self-stopping by local pressure effects. In contact with forming martensite the austenite will be under compression, but at a distance other austenite must be under tension to bring about the general balance of tension-compression stresses through the piece as a whole. This tends to make a banded intermix of martensite and austenite layers, with a spacing fixed by the stress variations. It is this banded intermix of austenite, martensite, and probably troostite which is commonly called martensite. Further, during a sustained cooling the outside of a piece, being cooler than the inside, tries to shrink more, and so the outside goes into tension, the inside into compression. The severity of these stresses depends on size and shape of piece. Their general effect is to promote martensite formation in the outside parts, and inhibit it in the inside.

Instead, then, of a simple mass law statement of the formation of martensite as a first order reaction, in the form $dM = -dA$, and $dA/dt = -RA$, where M = amount of martensite, A = amount of austenite, we must put $dA/dt = -R(A - bM \pm C)$. Recognizing that the amount of A , residual, equals $(A_0 - M)$, where A_0 is the original amount of austenite at time zero, we get $dM/dt = R(A_0 - M - bM \pm C)$. The term bM recognizes the local pressure increase to be proportional to the amount of martensite formed; the term $\pm C$ represents the regional pressures due to continued cooling. At steady temperature the term $\pm C$ vanishes. Integration gives

$$M = \left(\frac{A_0 \pm C}{1 + b} \right) \left(1 - e^{-(1+b)Rt} \right)$$

The final martensite is $\left(\frac{A_0 \pm C}{1 + b} \right)$; the retained austenite

is $\left(\frac{bA_0 \mp C}{1 + b} \right)$. We would not want to argue that these equations hold clear to the end of the action, but they characterize its beginning and give the nature if not the exact values of the

final results. Mathematically, the factor $(1+b)$ becomes a scaling factor on both the ordinate and abscissa of the curve of M versus t of martensite formation; the final amount of martensite formed is less than without the pressure effect in the ratio $1/(1+b)$; and the time to finish the reaction, or to reach a given relative stage of reaction, is less in the ratio $1/(1+b)$. And in proportion as formation of martensite is stopped, austenite is retained. Here we see why, also, oil and water quenches differ, through the operation of the factor $\pm C$ of regional pressures; and why continued cooling bringing in the factor $\pm C$, gives a different result than steady temperature.

Pressure effects can occur to reactions in the solid whenever volume changes accompany the reaction and the elastic limit and moduli of elasticity are high enough to permit the pressures to build up. The volume change austenite to pearlite is only half that of martensite formation, but when the martensite critical temperature is low enough, the immediately preceding pearlite formation can show also an austenite stabilization and retention. This happened in one of Davenport and Bain's steels.

Pressure effects are also important in the tempering of hardened steels, making the changes during heating up in the tempering bath much more rapid than the continuing changes after a piece comes to temperature in the bath—quite a time-saving scheme for practical heat treatment. Time in the tempering bath becomes important chiefly for relief of cooling stresses.

Hardened steels contain more or less of many structures—austenite, pearlite or sorbite, martensite, troostite, to name the four groupings which have fairly definite sets of properties. To find out how much a steel contains of each of these requires measurement simultaneously of at least four sets of properties—four unknowns require four equations to fix them. Until, then, systematic experiments are done wherein many samples of one steel are put through many heat treatments and stages of treatment, and at each stage at least four critical properties of the metal are measured simultaneously, we cannot solve the problem of the amounts of the various structures in the steel and the contributions of each to the final results. Such research work in heat treatment theory and practice remains to be done. We have made a start.

DISCUSSION

DR. JOHN JOHNSTON, *Chairman*

H. A. SCHWARTZ, (*Manager of Research, National Malleable and Steel Castings Co., Cleveland.*): A discussion of so generalized a paper as Upton's must of necessity be limited to certain "high spots" which happen to attract the attention and interest of the commentator.

The question of whether alpha and beta iron are, or are not, two

distinct phases has become, in the writer's mind, purely a matter of definition. If entities are to be separable from one another there must be some criterion by which they differ and since magnetism furnishes such a criterion there appears reason to consider them as distinct phases. On the other hand, the concept of phases and components arose out of the Gibbs investigations on heterogeneous equilibria. The conclusions which are now embodied in the Phase Rule imply a difference in free energy content between two phases of the same substance. From this viewpoint the coexistence of two phases in a one component system is impossible over a temperature range at constant pressure. The usual assumptions as to the character of the magnetic change near A_2 are then not in accord with the concept of a distinction of phase. When we contemplate the possibility of the existence of two forms of a pure substance differing in some detectable property but not in energy content, ambiguity is unavoidable. The fact that A_2 is not depressed by solutes points to a small difference at most of free energy level between the magnetic and nonmagnetic states.

In doubtful cases the present writer happens to prefer an interpretation in the light of the phase rule to any other, feeling it more consistent to leave that unaltered at the expense of redefining a phase rather than the reverse.

It appears that in Figs. 3 and 4 the author has not actually adhered to his cube root relation, for he draws a smooth curve through the observational values and not a straight line as he would have done had he had unbounded confidence in the cubic function. Drawing a smooth but empiric curve through points plotted as a function of the cube root of the rate of change of temperature with time yields no information not furnished by a similar curve in which the cooling or heating rate rather than its cube root is used as the abscissa. Either curve would intersect the line of zero cooling (or heating) rate at the same point. It may be suggested that the author would have more consistently adhered to his own thesis had he represented these plots by the best straight line which might be drawn through them.

The explanation of the occurrence of the stable or metastable systems during the cooling of iron-carbon alloys, substantiating the so-called Stufenregel of Tamman by calculation of the point at which a eutectic, or eutectoid, should form as a function of cooling rate is, to the present writer, the most interesting part of the paper. This interest is not materially decreased even if the cube root relationship should ultimately not be completely substantiated. (*Written Discussion.*)

R. F. MEHL, (*Director, Metals Research Laboratories, Professor of Metallurgy, Carnegie Institute of Technology, Pittsburgh.*): There is no subject that can be presented to a metallurgical audience more powerful in producing discussion than that of the de-

composition of austenite. As a matter of fact the great wealth of information that is now available on the subject and the very many scientific aspects of the problem make it a bit difficult to restrict a discussion of the general phases of the subject within any reasonable scope. This paper by Professor Upton is extremely thought-provoking. It has, as you may see, two major sections, the first concerning the nature of the α - β change in iron and the second, which is the main topic of the paper, on the decomposition of austenite. The discussion I have prepared concerns chiefly the second, but before entering into this part, I should like to comment briefly on the first part.

The way in which Professor Upton has drawn the Fe-Si diagram is very ingenious. I believe, however, there is little scientific justification for it. Evidently much of the argument must depend upon the point of view adopted considering the α - β change in iron. Whether this should be regarded as a true allotropic change or not is of course an argument which has grown hoary with age in metallurgical circles and about which there can still exist two classes of opinion. I am among the group which sees little reason for considering the change as a phase change. Let me illustrate by listing the arguments on this side of the question. The α - β change in iron is one which is shown by the change in the magnetization, the α phase losing its ferromagnetism on heating at the A_2 point. Accompanying this change there is also a change in heat content, for heat absorption is noted. This reaction, however, is a curious one. It is characterized by no change in density, by no change in a , the side of the unit body-centered cubic cell. There is no change in microstructure and accordingly there can be, and is, no possibility of altering the structure of iron or steel by heat treatment involving the use of the A_2 point. This change differs from other changes which are known to be true phase changes in a number of very important respects. The magnetism is not lost at constant temperature. It is lost over a range of temperatures beginning at a relatively low temperature and extending up to 800 degrees Cent. The accompanying heat change is likewise distributed over a range of temperature of about 20 degrees Cent. reaching a maximum at a point where the magnetization curve has the greatest slope. I shall show these data on the board drawn from experiments conducted in my laboratory by Dr. Walters. The change further differs in the fact that it shows no hysteresis. The A_2 point for a given sample of iron is at the same temperature for the rates of cooling varying from 15 to 250 degrees Cent. per minute heating or cooling. Evidently therefore we are not dealing with a heterogeneous reaction where one phase changes into another, for such reactions must take place at a constant temperature and must by the nature of heterogeneous reactions be characterized by hysteresis, that is, must be characterized by a time rate of change. Both of these factors are of importance in arguments which are intended to lead to an adoption of points of

view concerning the nature of the A_2 point. We may also characterize a true allotropic change by the behavior of the thermodynamic potential or free energy curves. Each phase will have its own thermodynamic potential or free energy curve, and these curves will not intersect over a range of temperature but at a particular temperature. This is perhaps as close a definition of the meaning of phase as we can attempt, and in view of the characteristics of the α - β transformation this transformation cannot be classed among true phase changes.

Accordingly, therefore, it should not be permissible to combine a curve of magnetic change with curves of true phase changes upon a constitutional diagram as shown in Fig. 2. We may argue the point somewhat obliquely by stating that no satisfactory combination of this sort has ever proved useful in the study of metal systems, but I should prefer to argue it more directly, namely, that the thermodynamic principles governing phase changes are well understood and unique; whereas no ordinary thermodynamic treatments can be advanced for governing the changes in magnetism, which are purely intra-atomic.

The iron-silicon system is, to be sure, a complicated one, and there remains a great deal to be done in studying the behavior of this system in its relationship to constitution. Many of the peculiarities come from the well known brittle constituent present in electrical sheet. This constituent is, however, a carbide and is therefore an impurity in the iron-silicon system. The peculiar behavior of the lattice dimensions in the range up to 20 per cent silicon, as shown by Phragmen and later by Jette, is at the moment entirely unexplained. Incidentally both these workers find that the side of the unit cube varies progressively through the field which in Fig. 2 is marked as a heterogeneous field. This would not be the case were the field truly heterogeneous. Heterogeneity is characterized at a constant temperature by unchanging values of the lattice parameters of the constituent phases throughout a heterogeneous field. Although the combination shown in Fig. 2 seems to simplify the constitution of iron-silicon alloys, I am inclined to believe that it is an oversimplification and scientifically not justifiable.

I might note that there is considerable justification for the definition of a phase as a unit cell which Professor Upton calls into question. A unit cell, of course, is not a concept but a physical entity determined directly by X-ray diffraction. The unit cell will distinguish between what Professor Upton calls order and disorder in the arrangement of atoms in a space lattice. For those beta solid solutions which are body-centered cubic the unit cell of the disordered lattice is simply an ordinary body-centered unit cell, but for the ordered lattice it is designated a lattice with a super-structure. The unit cell is again body-centered cubic but of the caesium chloride type in which the body-centered cubic cell is obtained by the interpenetration of two simple cubic lat-

tices, one composed of one sort of atoms and one composed of another. Accordingly the specification caesium chloride type gives a unit cell, different in this way from that when the specification is body-centered cubic alone.

Besides the general interest attached to Professor Upton's discussion of the decomposition of austenite which I believe to be quite new in many respects and to be very valuable, I find the discussion interesting because it is directly related with work which is going on at the moment in my own laboratory. The treatment of the results of Sato shown in Figs. 3, 4, and 5 is very interesting. I wonder if Professor Upton has any explanation of the importance of the cube root of the cooling or heating rates and straightening out of the critical points found. The method he uses, however, is a very convenient graphical method of evaluating true equilibrium temperatures even though the meaning of the cube root may not immediately be apparent. Sato's data are quite good data. It must be pointed out, however, that Sato's alloys were not especially pure, and I believe probably somewhat irregular with respect to purity. You will note from Sato's data that the silicon content occasionally was sufficiently high to account for a rather appreciable increase in the $A_{1,2}$ temperatures. According to Sato's analyses the percentage of silicon varies around an average of 0.10 per cent, lower in the hypoeutectoid alloys and higher in the hypereutectoid.

The method of preparing the steel is, as Professor Upton suggests, of great importance in determining the amount of undercooling which the alloy will exhibit and will therefore determine the curves shown in Fig. 7. Incidentally this treatment given in Fig. 7, I believe, is new, and I find it very illuminating. There are two studies in this general field now current in Pittsburgh, one is being performed in my laboratory and the other in the laboratory of Dr. Herty. The methods of attack used are somewhat different though they lead to similar results. We have adopted the dilatometer at least for the moment in studying the temperatures of transformation at varying cooling rates and also the rates of precipitation. The first slide shows the type of curve which is obtained on heating and cooling on 0.40 per cent carbon steel, prepared by Howard Scott of the Westinghouse Electric and Manufacturing Co. from hydrogen-purified iron, melted and alloyed with carbon and poured under hydrogen. The next slide shows the dilatometric cooling curves obtained from this sample at various rates of cooling from 1.3 degrees per minute to 50 degrees per minute. You will see that the $A_{1,2}$ point is greatly affected by the different rates, but that the A_1 point is not much affected. This is quite in conformity with Professor Upton's diagram. The next slide shows the same steel in which the temperature to which the material was heated before cooling was increased in four steps. Evidently the effect of heterogeneity of which Professor Upton speaks and of grain size plays a role here

for the temperature is lower the higher the steel had been heated. The next slide shows the steel prepared in similar way with 0.27 per cent carbon. The effect of heating at various temperatures in the austenite zone is very evident here also. The next slide illustrates the effect of what might be called metallurgical condition. The upper two curves show the position of the transformation temperatures for cooling rates of 2.7 degrees Fahr. per minute and 22 degrees Fahr. per minute. The steel was prepared by Dr. Herty and had been thoroughly killed with aluminum. It contained 0.25 per cent carbon. The lower two curves are similar but for the material before the addition of aluminum, therefore, for a wild steel. It can be seen that the amount of undercooling is less than for the killed steel. The next slide shows dilatometric curves taken on the wild steel from different temperature points within the austenite range. Apparently the wild steel is not so sensitive to the effect of heat treatment within the austenite range as the previous laboratory pure steels. (These curves will shortly be reproduced in a publication, and for this reason are not given here.—Ed. note.)

The technique adopted by Mr. McBride in Dr. Herty's laboratory is somewhat different from this. In this case the rate of precipitation is studied using the same technique used by Bain and Davenport. The next slide is a summary of the results. We have plotted here the decrease of super-cooling as abscissae and the percentage of ferrite per minute divided by the grain surface per cubic inch as ordinate. The selection of this unit on the ordinate you will see eliminates the effect of grain size. It was observed in these studies that the ferrite precipitates only at the grain boundaries. Steel Z is an aluminum-killed steel with 0.41 per cent carbon. Steel D is a similar steel with 0.44 per cent carbon. Steel Y is a silicon-killed steel with 0.41 per cent carbon. Steel AD is a silicon-killed steel with 0.40 per cent carbon. All of these data show what might be described as the effect of the metallurgical condition of the steel. Apart from this, ordinary analysis chiefly with respect to carbon is of great importance in determining the rate of ferrite precipitation. It is evident to many of you here that the results of these studies should be of interest in the heat treatment of plates, forgings, and steel castings, especially with respect to normalizing and the slower rates of quenching.

Professor Upton has followed Bain and others in drawing analogies between the reaction for the formation of pearlite and ordinary chemical reactions with respect to rates. While it is quite true that formal compliance can be obtained in this way, I think there are some distinct advantages in analyzing rates such as this from a standpoint of the known mechanics of the reaction. The formation of pearlite from austenite proceeds by the formation of nuclei which grow forming essentially spheroids. The spheroidal nature of the pearlite or troostite-pearlite patches is quite evident in many of Bain's photomicrographs. Evidently if we can set up

equations which express the rate of reaction in terms of the known process for the formation of pearlite from austenite, our equations will have a specific physical meaning. This has been done by Göler and Sachs.* I thought it might interest you to see this and to have the advantages of the method pointed out. Bain's curves may be analyzed according to this method. In fact Wever's curves which are given in the diagram are much similar to those of Bain's Fig. 1. It will be noted from the derivation of the formula from which the curves were drawn that one factor has been left totally unaccounted for, namely, the decrease in active growing surface of the spheroids by reason of spheroids growing together and thus decreasing the growing surface. I have myself attempted to complete the equation and to draw the full curve, but the methods of analysis are very complicated involving some rather uncertain statistical considerations.

These processes consist in formation of nuclei and growth of nuclei. Assume a constant rate of nuclei formation and a constant rate of growth—constant growth velocity.

The volume Φ of a cubical grain at time t whose nucleus formed at time T is:

$$\Phi = 8 w^3 (t - T)^3 \quad (1)$$

Here we assume the grain to form a cube, and the rate of growth of each of the six faces is w .

For the growth of spheroids $8 w^3$ is replaced by $\frac{4\pi}{3} w^3$

Let $u(T)$ be the fraction of the volume unit still not crystallized at any time instant.

At time T in this volume $u(T)$, dN crystals form, so that

$$dN = c \cdot u(T) dT \quad (2)$$

These have at time t , according to equation (1) the volumes:

$$\Phi = c \cdot u(T) dT 8w^3 (t - T)^3 \quad (3)$$

1 — $u(t)$ = crystallized volume at time t . This can be obtained by integrating equation (3) for times between zero and t .

This gives the equation

$$u(t) = 1 - 8cw^3 \int_0^t u(T) (t - T)^3 dT \quad (4)$$

$$u(t) = 1 - \frac{4\pi}{3} cw^3 \int_0^t u(T) (t - T)^3 dT$$

Substituting values for c and w then gives the curves shown in Fig. 1. You will see that the upper portions of Sachs' curves do not correspond with the experimental curves because the factor of decreasing surface area has been left out of consideration. It appears to me that with a full derivation of the reaction curve in such terms as these the possibility exists, at least, of analyzing the reaction into its component parts, namely, the single factor of rate of nuclei formation and the other single factor of crystallization velocity may be studied. If metals were not opaque, the analysis of reactions of this sort in this way would be much simpler!

Evidently both factors of nuclei formation and crystal growth

*F. v. Göler and G. Sachs, "Zur Kinetik von Kristallisationsvorgängen," Zeitschrift für Physik, Vol. 77, 1932, p. 281-286.

will be affected by diffusion and Professor Upton properly points out the importance of this. I should think it would be very interesting to attempt some quantitative explanation for the effects of manganese, nickel, and other alloying elements on rates of reaction. It is barely possible that the effect of these elements on rates of diffusion might go a long way to explain the retardation in the reaction which alloying elements exert.

While it is quite legitimate to discuss the pearlite reaction on this basis since it proceeds in such an orthodox fashion as indicated, the martensite reaction is in a class by itself. Here the reaction does not proceed by the formation of nuclei and their subsequent growth but proceeds by a shift in the lattice, a shift in the nature of a shear parallel to the octahedral planes of the austenite. The full mechanics of this shift are now well understood. Both the tetragonal phase and the ferrite possess unique orientations with respect to the mother austenite. Each orientation of austenite is capable of producing 24 new orientations of either the tetragonal phase or the ferrite. It has been generally assumed, especially by German workers, that the martensite reaction differs from a pearlite reaction by the fact that the pearlite reaction generates no specific orientations in the pearlite. How-

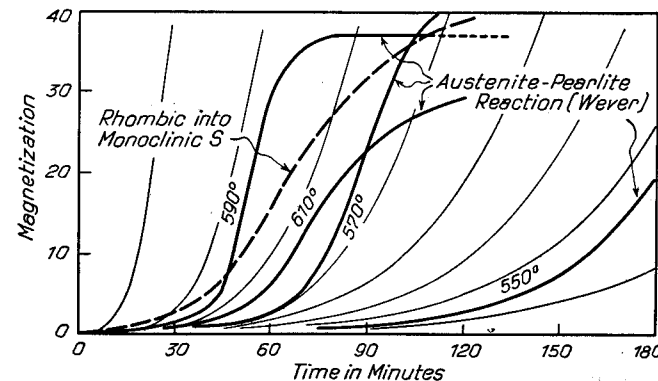


Fig. 1

ever, this question has been restudied in our laboratory this year, and it has been found very recently that the ferrite in pearlite possesses an orientation which it inherits directly from the original austenite much as the tetragonal phase and the ferrite in martensite inherit their orientations from the original austenite. There is a difference, however, between the two reactions from a purely crystallographic point of view apart from their characteristic reaction rates, namely, the orientations possessed by ferrite in pearlite with respect to the mother austenite are not those displayed by the tetragonal phase and ferrite in martensite with respect to the austenite. Accordingly we have two different crystallographic mechanisms for the decomposition of austenite.

Whether such crystallographic points of view will ultimately be of much use in the analysis of rates, of course still remains to be seen. (*Written Discussion.*)

MR. DOAN: It is difficult for me to conceive of a paper better suited to an audience interested in the heat treatment of steel than the one just presented by Professor Upton. The interpretations of Professor Upton are quite worthy of the experimental results, especially the classical ones of Davenport and Bain, on which they are based. In view of the limited time available for discussion, I shall attempt to speak upon several points which are of interest to many of you rather than to discuss matters of greater detail.

In the first place, I agree with Dr. Mehl that the X-ray method should be given credit with showing a difference between the beta and the beta-prime phases in brass. The presence of a superlattice in the latter phase seems to me adequate indication of a separate phase.

The method of locating the true Ac and Ar points which Professor Upton has used is, I think, also worthy of attention because of the very great importance of these temperatures. His method of determining the exact eutectoid composition also is a valuable one.

In connection with Fig. 7 and the discussion of these quasi-equilibrium diagrams, I must admit that it is difficult for me to visualize an austenite solid solution precipitating a constituent in which it is not saturated, that is, when it is saturated actually in another constituent—for this is what we are led to believe if this diagram is interpreted according to Professor Upton. This follows from Professor Upton's statement that pearlite does not supercool as much as either proeutectoid alpha iron or Fe₃C.

The author has also made quite clear to us again the effect of holding a steel in the austenite region long enough for the composition to become practically uniform in all parts of each grain. He has pointed out again the difference in composition of austenite held at different temperatures and also the effect of grain size in determining how long it is necessary to hold the steel in the austenite region. These well known requirements of heat treatment are clearly treated by the author in terms of the supercooling lag of the martensite transformation, shown also at the lower right-hand part of Fig. 7.

I am especially pleased with the author's treatment of the rate of transformation of austenite at various sub-critical temperatures, as one influenced by (a) the absolute temperature and (b) the distance of the reaction temperature from equilibrium temperature. This is a perfectly fundamental viewpoint and its value is revealed in the good correlation obtained in the author's calculations, and his graphs (Figs. 12 and 13).

Concerning the view that the hardness of martensite is due alone to the fineness of grain size, it seems to me that the author is on rather dangerous ground in his conclusion. Sachs' relationship of grain size and hardness shows a large effect of grain size as we approach a single crystal, but very little increased hardening when the grain size becomes finer. It may be that hardness is accompanied by a fine size; but in view of the modern knowledge of age hardening, it seems to me doubtful that the hardness of martensite is due to grain size alone.

MR. UPTON: Rarely does discussion add so much to a paper as it did in this case. Particular thanks are due to Dr. Mehl.

The author does not know why pearlite should supercool less, or form faster, than its components do alone. Such seems simply to be the fact. Dr. Johnston states that the driving force to form nuclei varies as $\sqrt[3]{T_c - T}$, which would give basis for the "law of supercooling" suggested by the author purely as empirical. Mr. Cowle confirms the finding of 720 degrees Cent. for Ac. Dr. Schwartz will find that while carbon is too slightly soluble in α or β iron to affect the magnetic change point, other solutes, for example Ni, Mn, Si, P, do considerably affect it. The author's Fig. 2 does not violate the phase rule.

Between the author and Dr. Mehl discussion might be endless. The specification of a phase as a space lattice pattern seems incomplete; the nature of the atoms occupying the points of the lattice would seem also important. Atoms of α and β iron are not alike. The curve of "orderliness" versus temperature for β and β prime in brasses is exactly like, in form, the curve of permeability (in strong fields) versus temperature in α and β iron, except that the high and low temperature ends are transposed. Magnetic testing has fallen into a habit of taking magnetometer readings "on the fly" while temperature is rising or falling. While this may be justifiable from the way magnetism varies with temperature in an element of material (and that is arguable), it involves, near the magnetic change point, trouble from the fact that different parts of the test piece are at different temperatures, and the magnetometer gives an integrated reading of all the parts. Hence the procedure causes a rounding of what might, perhaps would, be sharp corners on the permeability-temperature curves. Magnetism often shows temperature hysteresis; look at the so-called irreversible nickel and manganese steels, or the magnetic change of Fe₃C.

The nickel alloys of iron may illustrate another possibility also. We cannot make artificially the structures of meteorites—it takes too long. The equilibria of the Fe-Si alloys are likewise slow at even medium Si; hence they may act as if the line F B (Fig. 2) extended downward to the right, unless they are given a very long anneal at the temperature of test. This would explain the low temperature X-ray findings.

The Göler and Sachs equations for reactions in the solid are of a form that can characterize only the beginning of a reaction in the solid. But they are on the right track. Two time rates are involved, that of formation of nuclei and that of growth upon the nuclei. Gaseous reactions would lack, perhaps, one of these. In the case of pearlite formation, the nucleus formation rate is slow compared to the growth rate; in the case of martensite formation, the nucleus formation rate is high compared even to the high growth rate. Further, in the case of martensite, growth is much more rapid in two dimensions than in the third. The fact that the form of the curve of pearlite formation versus time is independent of the temperature shows that temperature affects alike, by both the functions of T abs. and $(T-T_c)$, the nucleus formation rate and the growth rate—a most important point for the final mathematical statement of our problem.

THE P-F CHARACTERISTIC OF STEEL

by B. F. SHEPHERD

Manager, Rock Drill Division, Ingersoll-Rand Company

FOR many years steel makers and steel users have expressed preferences for certain brands of steel claiming superiority due to some inherent property vaguely described as “body,” “timbre,” “personality,” etc. Considerable controversy has arisen as to whether such major differences in quality could exist between steels of apparently similar chemical composition.

In 1922 McQuaid and Ehn found that certain heats of steel had a tendency to harden in spots after being carburized and quenched from the hardening temperature. They were able to isolate such heats of steel by the use of the McQuaid-Ehn test, which consisted in noting the condition of the hypereutectoid zone after carburizing at a high temperature. This immediately resulted in a demand for the so-called “normal” type and since that time there has been very little “abnormal” carburizing steel manufactured.

At some later date it was found that the grain-size of the hypereutectoid zone of steel given the McQuaid-Ehn test gave an index of the response of that steel to heat treating operations. In general, the so-called coarse-grained steels hardened harder and deeper, and had much lower impact strength and better machinability.

In 1926 we began distinguishing the differences between tool steel by means of the Shepherd Hardenability test. This test was made at the ordinary range of hardening temperatures but elevated temperatures were included later. A considerable difference was found in the behavior of various heats (of the same hardenability at 1420 degrees Fahr.) after quenching from elevated temperatures. This difference consisted in a change in the hardenability and also in the martensitic fracture grain-size.

Classification of fracture grain-size was extremely difficult until a set of martensitic fracture grain standards was developed. It was then possible to describe fracture grain-size tests numerically. These standards are shown in Fig. 1 with corresponding microstructures in Fig. 2.

The classification of fractures requires a great deal of personal

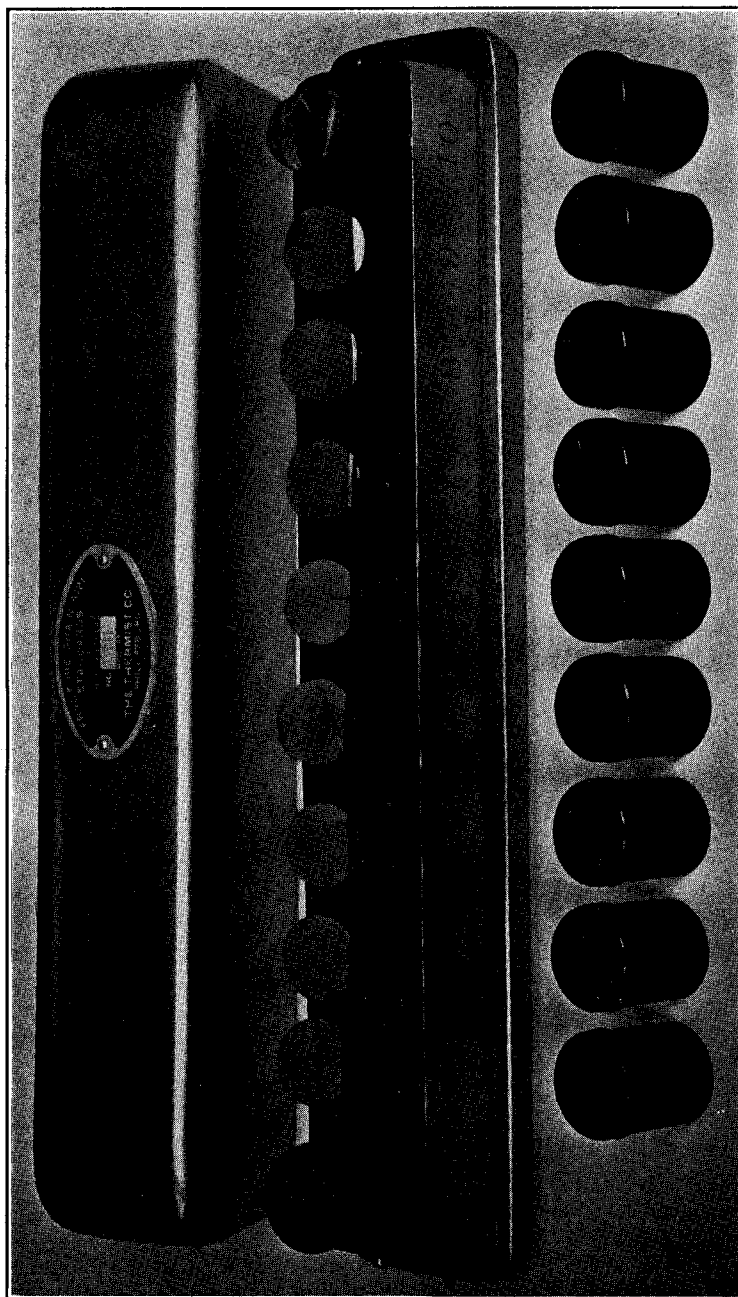


Fig. 1—Set of Martensitic Fracture Grain Standards Showing Protective Covers Removed from Specimens 2 to 10.

skill. Even though a set of fracture standards is available, proper technique must be used in interpreting values. The following suggestions are made as an aid to such interpretations:

1. Any fracture determinations should be an agreement of two qualified observers. Disagreement by more than $\frac{1}{2}$ fracture number shall result in arbitration by a third observer and the apparent true fracture number recorded.

2. To assist in eliminating personal error, classifications are made to $\frac{1}{4}$ number.

3. Observers are to be qualified by passing the following tests on a set of approximately twenty samples of various fracture classifications.

- Ability to check their own classifications on different days to within $\frac{1}{4}$ fracture number on 90 per cent of the samples.
- Apparent true values should be established on approximately twenty samples by two or more observers passing Test No. 1. To qualify, at least 80 per cent of these samples must be classified within $\frac{1}{4}$ fracture number.

A representative qualification test is shown in Table I.

TABLE I
QUALIFICATION TEST FOR FRACTURE OBSERVER

Sample No.	Apparent True Fracture No.	Observer Variation from Apparent True Number			
		A	B	C	D
1	8	$+\frac{1}{4}$	0	$+\frac{1}{4}$	-1
2	$4\frac{1}{2}$	0	0	0	-1
3	$7\frac{1}{2}$	0	$+\frac{1}{4}$	0	-1
4	$8\frac{1}{4}$	0	-1	0	$-2\frac{1}{4}$
5	$3\frac{1}{2}$	$+\frac{1}{4}$	$+\frac{1}{4}$	$-\frac{1}{4}$	$-\frac{1}{4}$
6	$8\frac{1}{2}$	0	$+\frac{1}{4}$	0	$-\frac{1}{2}$
7	5	0	0	$+\frac{1}{4}$	0
8	$8\frac{1}{4}$	0	0	$+\frac{1}{2}$	0
9	3	0	$+\frac{1}{4}$	0	0
10	$8\frac{3}{4}$	$-\frac{1}{4}$	$-\frac{1}{4}$	$+\frac{1}{4}$	0
11	$8\frac{3}{4}$	$+\frac{1}{4}$	$-\frac{1}{4}$	0	$-\frac{1}{4}$
12	$7\frac{1}{2}$	0	0	$+\frac{1}{4}$	-2
13	$4\frac{1}{4}$	0	0	$+\frac{1}{2}$	$+\frac{1}{4}$
14	$8\frac{3}{4}$	0	$-\frac{1}{4}$	$+\frac{1}{4}$	$-\frac{1}{2}$
15	$3\frac{3}{4}$	$+\frac{1}{4}$	0	$-\frac{1}{4}$	0
16	$8\frac{1}{4}$	0	$-\frac{3}{4}$	$-\frac{1}{4}$	$-\frac{3}{4}$
17	$3\frac{3}{4}$	0	0	$-\frac{1}{4}$	$+\frac{1}{2}$
18	$8\frac{3}{4}$	$-\frac{1}{4}$	0	$-\frac{1}{4}$	$-\frac{1}{4}$
19	8	0	$+\frac{1}{2}$	0	$+\frac{1}{2}$
20	3	0	0	$-\frac{1}{4}$	0
Grade		100	85	90	50

PENETRATION (P) — FRACTURE (F) TEST METHOD

1. Preparation of Test Pieces

(a) It is most important to obtain metal that is representative of the heat of material to be tested. Therefore, stock whose dia-

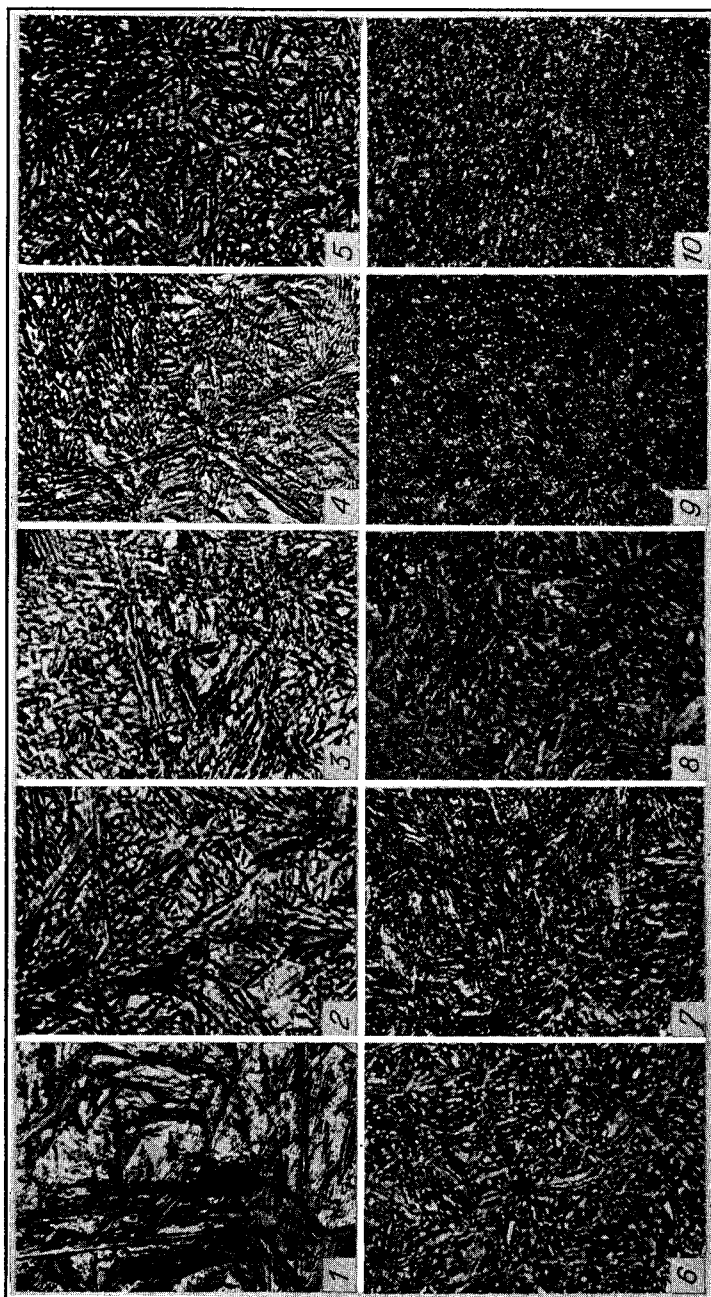


Fig. 2—Photomicrographs of the Structures of Fracture Grain Standard's Shown in Fig. 1.

meter is greater than $1\frac{1}{4}$ inch is reduced in size to 1-inch round (approximately) by suitable forging. Approximately 14 inches of material is required in lengths not less than 3 inches. When possible, the original marking is allowed to remain untouched in the forging operation, thus avoiding mixed samples.

(b) The sample material may be annealed before machining (1325 degrees Fahr. air or 1450 degrees Fahr. furnace cool, as convenient).

(c) Four specimens are machined $\frac{3}{4}$ -inch round, plus or minus 0.001 inch and 3 inches long. Each piece is stamped with a suitable code or number for identification. The pieces are also stamped on both ends, 1450, 1500, 1550 and 1600 respectively, one number to each piece.

2. Normalizing

All samples are given an oil-quench from 1600 degrees Fahr. after holding the pieces at temperature a minimum of 40 minutes after heating through.

3. Hardening

(a) HEATING—An electric furnace should be used for heating. The heating rate should be similar to or the equivalent of that obtained in a 11 KW top and bottom element electric furnace

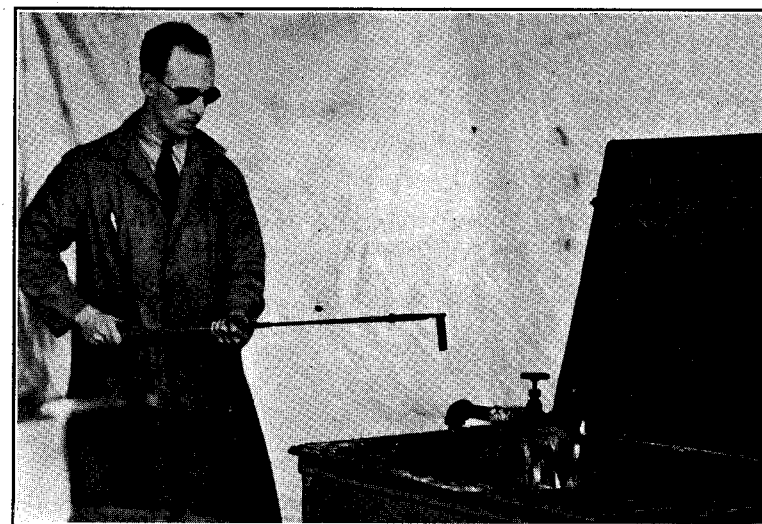


Fig. 3—Showing Method of Quenching Specimens in a Standard Vertical Jig.

with a hearth size of 12 by 30 inches, and not more than four samples shall be heated at one time. The samples shall be placed erect in the furnace, and so located that all samples are heated at the same rate. Temperatures of 1450, 1500, 1550, and 1600 degrees Fahr., respectively, shall be used, the samples having been

so identified previously. The samples are in the furnace for 30 minutes for each temperature.

Accurate control of the time-cycle and temperature must be maintained. This is extremely important because penetration of hardness and fracture grain-size tend to increase with time and temperature on all except very stable heats.

(b) **QUENCHING**—The samples shall be quenched in a standard vertical jig (2-inch pipe overflow, Fig. 3), flushed with 10 per cent brine at room temperature.

Fig. 3 shows a specimen in the tongs being placed in the spray. The tongs have a shoulder which enables the specimen to be placed easily in the center of the spray. The P-F test section (Fig. 5) shows the uniformity of distribution of hardness along the length. The soft spot has been caused by the tongs.

4. Fracture Grain-Size

(a) **BREAKING**—The samples are notched halfway between ends with a thin ($\frac{1}{16}$ to $\frac{3}{32}$ inch) friction wheel, to a depth not over $\frac{1}{16}$ inch. The fracture is obtained preferably by transverse impact because transverse static loading to rupture tends to pull or otherwise distort the grain.

(b) The grain-size is determined by comparing and matching the case or hardened zone of one of the broken halves with the fracture grain-size standards, and shall be expressed as the number or quarter number nearest the corresponding standard sample. Determination of the grain-size number is made according to the standard method.

5. Penetration

The penetration tests are made on the second half of the hardened and fractured sample. A smooth cross section surface is prepared not less than 1 inch from the original end of the hardened

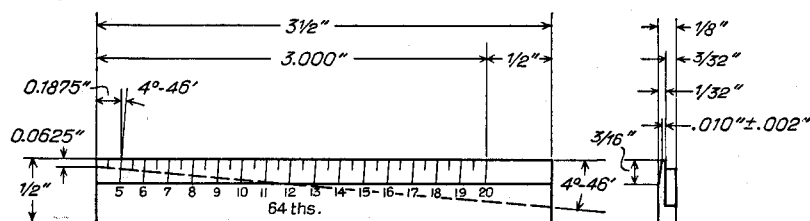


Fig. 4—Gage for Measuring P-Value.

sample. Final finish on the cross section surface is on an "0" or finer abrasive. The polished surface is etched for 3 minutes in a 50-50 hydrochloric acid and water at 180 degrees Fahr.

The hardness penetration is measured, macroscopically, to the nearest half of $\frac{1}{64}$ inch and shall be expressed as the numerator of such a fraction. A convenient scale is shown in Fig. 4.

6. Designation

The P-F characteristic is recorded as 8 numbers. The first four numbers represent the penetration in $\frac{1}{64}$ ths of an inch, and the last four numbers the fracture grain-size. The first number of each group represents the P-F result for a temperature of 1450 degrees Fahr., the second number of each group represents the P-F for a temperature of 1500 degrees Fahr. and so on.

The characteristic then appears in the following form:

5. 6. 8. 12. / 9. 8. 6. 4.

When all the results do not concur with whole numbers, the characteristic may appear as follows:

5½. 6. 8½. 13. / 8¾. 8¼. 7½. 5.

In developing the test method electric furnaces were compared against the lead pot method of heating. The following typical

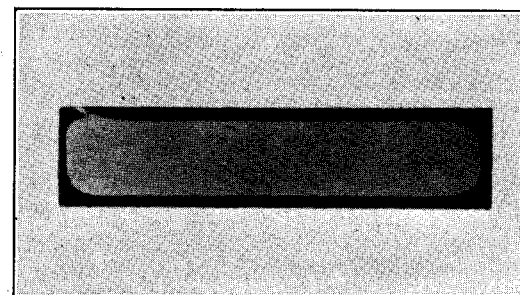


Fig. 5—P-F Test Section Showing the Uniformity of Distribution of Hardness Along the Length. The Soft Spot Was Caused by the Tongs.

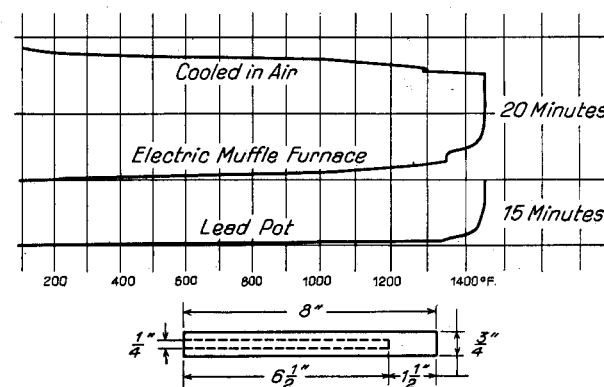


Fig. 6—Comparison of Rate of Heating a $\frac{3}{4}$ -Inch Round Specimen of Electric Furnace 1.05 Per Cent Carbon Tool Steel in an Electric Muffle Furnace and Lead Pot.

heating curves (Fig. 6) are characteristic of the results. A lead pot takes almost as long to heat these small samples to tempera-

ture as an electric furnace. The initial heating is very rapid and if a piece $\frac{3}{4}$ -inch round was quenched after five minutes in the lead pot it would harden, whereas in the electric furnace it would be just about in the transformation range.

In comparing oil-treated versus annealed samples it was observed that the oil-treated samples apparently heated much more rapidly than the annealed samples. Investigation with test specimens showed that this was due to a lowering of the critical point on the oil-treated samples as shown in Fig. 7. Apparently the carbides which have been made very small and uniform by the oil-treating operation are much more readily soluble. For a long, slow heating the critical points are the same. Critical point de-

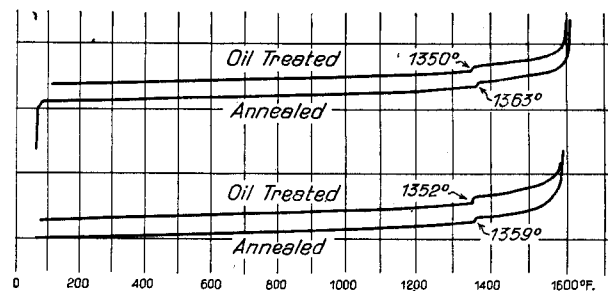


Fig. 7—Influence of Structure on the Critical Point of $\frac{3}{4}$ -Inch Round 1.05 Per Cent Carbon Tool Steel.

terminations which have been made by heating rapidly and have not been accompanied by a statement as to the previous microstructure, should not be given credit for strict accuracy.

The P and F values should not be considered separately when considering the P-F characteristic. In order to illustrate this, four heats are shown in Table II. On the first two heats, the P characteristic is practically the same, while there is a marked difference in the F characteristic. On the second and third heats the F characteristic is about the same but there is a marked difference in the P characteristic. The fourth heat has a very stable but coarse F characteristic.

It is not always desirable to give the full P-F characteristic value for a heat in discussion, and the heats are occasionally referred to as types. For tool steel we used arbitrary classifications as follows:

Type 1 was used to designate a steel which was stable or insensitive to increase in hardening temperature, both with regard to P and F characteristic and Type 3 as being directly opposite. Type 2 was intermediate.

Other types are classified as fractions. A Type 2-1, for example, is intermediate in P characteristic and stable with regard to its F characteristic; whereas a Type 1-3 would be stable with regard to penetration and very unstable with regard to fracture.

TABLE II
DIFFERENCE IN P AND F VALUES BETWEEN THREE BASIC OPEN-HEARTH HEATS

Heat No.	Penetration				Fracture			
	1450	1500	1550	1600	1450	1500	1550	1600
1D091	8	11	24	24	8 $\frac{1}{4}$	7 $\frac{1}{4}$	3 $\frac{3}{4}$	3 $\frac{1}{2}$
30420	9	11	24	24	8 $\frac{1}{4}$	7 $\frac{1}{4}$	6 $\frac{1}{4}$	5 $\frac{1}{2}$
5D330	7	8	9	13	8 $\frac{1}{4}$	7 $\frac{3}{4}$	7 $\frac{1}{4}$	5 $\frac{1}{2}$
CPX	12	16	16	24	4 $\frac{1}{2}$	4 $\frac{1}{2}$	4 $\frac{1}{2}$	4 $\frac{1}{4}$

Heat No.	Chemical Analysis				
	C	Mn	P	S	Si
1D091	1.01	0.27	0.015	0.021	0.15
30420	0.98	0.29	0.016	0.028	0.20
5D330	1.06	0.27	0.015	0.020	0.15
CPX	0.80	0.31	0.023	0.017	0.15

Type	P	F
1.	$\frac{3}{8}$ inch or less change.	Not more than one number.
2.	More than $\frac{3}{8}$ inch, but does not harden through.	More than one (1) number; less than two (2) numbers.
3.	Hardens through at 1600 Fahr.	More than 1 $\frac{3}{4}$ number change.

Incidentally, when a number of heats of various types were compared with the yield of finished bars passing fracture inspection, all of the points fell on the line; Type 1 having a high yield and Type 3 a comparatively low yield. Table III illustrates P-F types in electric furnace 1.05 per cent carbon steel based on the differences between 1450 and 1600 degrees Fahr. values.

TABLE III
P-F TYPES IN ELECTRIC FURNACE 1.05 PER CENT CARBON TOOL STEEL

Heat No.	Type	Penetration				Fracture			
		1450	1500	1550	1600	1450	1500	1550	1600
H5194	1	6	6	7	8	9	9	9	8 $\frac{1}{2}$
6101	1/3	5	5 $\frac{1}{2}$	6	7 $\frac{1}{2}$	9	8 $\frac{1}{2}$	6 $\frac{1}{4}$	4 $\frac{1}{2}$
H4705	2/1	6	7	8	12	8 $\frac{3}{4}$	8 $\frac{1}{2}$	8 $\frac{1}{4}$	8 $\frac{1}{4}$
X6892	2	5 $\frac{1}{2}$	6 $\frac{1}{2}$	8	14	8 $\frac{3}{4}$	8 $\frac{1}{2}$	8 $\frac{1}{4}$	7 $\frac{1}{4}$
H5183	2/3	5	5 $\frac{1}{2}$	7	13	8 $\frac{3}{4}$	8 $\frac{1}{2}$	8 $\frac{1}{4}$	5
29983	3	7	8	9	24	8 $\frac{1}{2}$	8 $\frac{1}{4}$	6 $\frac{1}{4}$	4
W5294	3	7	8	10	24	8 $\frac{3}{4}$	8 $\frac{1}{4}$	7	6 $\frac{1}{4}$

Heat No.	Type	Chemical Analysis						
		C	Mn	P	S	Si	Ni	Cr
H5194	1	1.07	0.21	0.010	0.015	0.17	0.09	0.03
6101	1/3	1.07	0.20	0.013	0.013	0.20	0.025	0.01
H4705	2/1	1.05	0.21	0.010	0.014	0.19	0.10	0.05
X6892	2	1.06	0.23	0.014	0.013	0.25	0.09	0.03
H5183	2/3	1.05	0.25	0.011	0.017	0.17	0.08	0.02
29983	3	1.06	0.07	0.011	0.011	0.14	0.06	0.05
W5294	3	1.05	0.21	0.011	0.015	0.24	0.06	0.03

In standardizing the test, heating times of 30, 45 and 60 minutes were used with typical results shown in Table IV. It is interesting to note that Specification 76 is so stable that it makes practically no difference in the penetration or fracture value, whether 1450 or 1600 degrees Fahr. is used for hardening.

TABLE IV
EFFECT OF HEATING TIME ON PENETRATION-FRACTURE VALUES OF
SEVERAL TYPES OF STEEL

Steel	Heating Time Minutes	Penetration Degrees Fahr.				Fracture Degrees Fahr.			
		1450	1500	1550	1600	1450	1500	1550	1600
ED	30	5	5½	6	6½	9	8¾	8½	8¼
Type 1	45	5	5½	6½	7	8¾	8½	8¼	8¼
Spec 1	60	5	5½	7	7½	8¾	8½	8	7¾
BM	30	7	8	9	17	9	8½	8	7½
Type 2	45	7	8	9½	T	8¾	8½	8	6½
Spec 1	60	7	8	10	T	8¾	8½	7½	6¼
DN	30	7	9	13	T	9	8¾	8¼	6½
Type 3	45	7½	10	14	24	8¾	8¼	7¾	5¾
Spec 1	60	7½	11	24	24	8¾	8	7¼	5½
2D	30	7½	9	12	24	8½	8	7½	6½
Spec 332	45	7½	9	14	24	8½	8	7	5¾
Type 3	60	8	9½	16	24	8½	8	6½	5½
2H	30	14	14	14	15	8	8	7¾	7¾
Type 1	45	14	14	15	15	8	8	7¾	7¾
Spec 76	60	14	15	15	15	8	8	7¾	7¾

Our previous tests and experience with regard to hardenability test specimens would indicate that the effect of spray or hand agitation would have little effect upon the results. This was borne out by the tests which were made on various types of steel as shown in Table V.

TABLE V
EFFECT OF SPRAY—S—IMMERSION VS. HAND—H—AGITATION IN QUENCHING
PENETRATION-FRACTURE TEST PIECES OF ELECTRIC FURNACE
1.05 PER CENT CARBON TOOL STEEL

Heat No.	Quench	Type	Penetration Degrees Fahr.				Fracture Degrees Fahr.			
			1450	1500	1550	1600	1450	1500	1550	1600
30127	H	1	5			6	9			8½
30127	S		5			6	9			8½
W5907	H	2	7			17	9			7½
W5907	S		7			17½	8¾			7¾
W5294	H	3	7			24	8¾			6¼
W5294	S		7			24	8¾			6
5760	H	1/3	5	5½	6½	7	8¼	7¾	5	4½
5760	S		5	5½	6½	7½	8½	7¾	5	4½
5877	H	2/3	5½	6	8	10	8½	8	5½	4¾
5877	S		5½	6	8	10	8½	8	5½	4¾
H4706	H	2/1	5			9	9			8¼
H4706	S		5			9	8¾			8

Segregation has an influence on hardenability. We find dumb-bell fractures in hardened test pieces occasionally and these are characteristic of the heat, showing a center condition which increases the depth of hardening. A standard ¾-inch slab test piece and two test pieces from different heats showing varying degrees of dumb-bell fracture are shown in Fig. 8.

Hardenability is a heat characteristic. A heat of electric fur-

nace steel was explored by taking the first, middle, and last ingots and making a Shepherd hardenability test from the top, middle and bottom of these ingots. Fig. 9 is self-explanatory.

An investigation of some defective steel showed that a reaction had taken place in the ladle which had progressed sufficiently far during the casting of the last ingot to cause a considerable difference in the billet tests, the etched bars and the P-F characteristic.

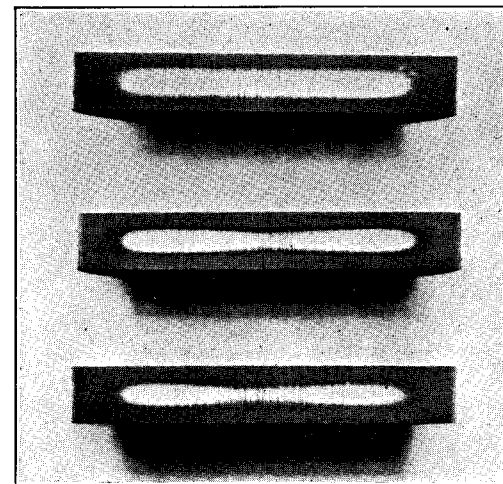


Fig. 8—A Standard ¾-Inch Slab Test Piece and Two Test Pieces from Different Heats Showing Varying Degrees of Dumb-bell Fracture.

This indicated that the P-F characteristic is due to the melting reactions and not to the ingot size, temperature of pouring, etc. Figs. 10, 11 and 12 are self-explanatory.

An electric furnace heat and an open-hearth heat were surveyed to determine whether the P-F characteristic was an inherent property of the heat itself. Tables VI and VII, Figs. 13 and 14, show that this characteristic is a function of the heat. It is evident, however, from Figs. 10, 11 and 12 that tests on both the first and the last ingot cast are necessary for an assurance of heat uniformity.

Hardness surveys are often made of specimens and the results assumed to be independent of the prehardening structure and typical of steel of that particular chemical analysis. The following graphs, Figs. 15, 16, 17 and 18, show that the P-F characteristic and preliminary structure have a great influence upon the shape of these curves.

TABLE VI

EXPLORATION OF HEAT 31196—SPEC. 1 FOR UNIFORMITY OF PENETRATION-FRACTURE CHARACTERISTIC. HEAT WAS FROM 8-TON BASIC ELECTRIC FURNACE TOP-POURED IN 15" SQUARE INGOTS

Penetration-Fracture Test Made on 1" Bars Hammered from Section of 4" Billet from Top of Ingot								
Results Given are Deviation from Average Results								
Ingot No. Average	Penetration Penetration in 64ths Degrees Fahr.				Fracture Fracture Grain Size Degrees Fahr.			
	1450 6	1500 6½	1550 7	1600 10	1450 8¾	1500 8½	1550 7	1600 5¾
1	0	0	0	0	0	0	¼	0
2	0	0	0	0	0	¼	¼	¼
3	0	0	½	0	0	¼	0	0
4	0	0	0	0	0	0	¼	½
5	0	0	0	0	0	0	0	0
6	This Ingot Scrapped in Processing							
7	0	0	½	0	0	¼	0	0
8	0	0	½	1	0	0	¼	0
Chemical Analysis of Heat								
C	Mn	P	S	Si	Ni	Cr		
1.06	0.19	0.010	0.014	0.26	0.05	0.05		

Material furnished courtesy Carpenter Steel Co.

TABLE VII

EXPLORATION OF HEAT 1D006—SPEC. 332 FOR UNIFORMITY OF P-F CHARACTERISTIC. HEAT WAS FROM 100-TON BASIC OPEN-HEARTH TOP-POURED IN 24" CORRUGATED UNITS

P-F Test Made on 1" Bars Hammered from Section of 4" Billet from top of Ingot												
Results Given are Deviation from Average Results												
Ingot No. Average	Penetration Penetration in 64ths				Fracture Fracture Grain Size				Chemical Analysis			
	1450 7½	1500 9	1550 12	1600 24	1450 8¾	1500 8½	1550 6½	1600 5½	C	Mn	Si	
D1	0	0	-2	0	0	0	+¾	+¼	-.02	-.01	-.01	
D3	0	0	0	*0	0	0	+¾	+¼	-.01	-.01	0	
D4	0	-½	+3	0	0	+¼	0	0	-.01	-.01	+.01	
D5	0	-½	-1	0	0	0	0	+¾	0	0	0	
D6	0	0	0	0	0	+¼	0	0	-.02	0	+.01	
D7	0	-½	-2	0	0	0	1	-¼	-.02	0	+.01	
D8	+½	0	0	0	0	+¼	-¼	-¼	-.01	0	0	
D9	0	-½	-1	0	0	+¼	-½	-½	+.01	0	0	
D10	+½	0	-1	0	0	0	½	-¼	-.01	.01	-.02	
D11	0	0	-1	0	-¼	-¼	-¼	-¼	0	.01	0	
D12	0	0	0	0	0	0	-¼	0	0	0	0	
D13	+½	0	-1	0	0	0	+¾	+¼	-.01	0	0	
D14	0	0	-1	0	0	0	+½	+½	0	0	-.01	
D15	+½	0	-1	0	0	-½	+1	+¼	0	0	0	
D16	0	0	+1	T	0	0	-1	-½	0	0	0	
D17	+½	0	0	0	0	-¼	0	-½	+.01	0	-.01	
D28	+½	0	+1	0	-¼	-¼	0	0	0	0	-.02	
D29	+½	0	+3	0	-¼	-¼	-¾	0	0	0	0	
D30	+½	0	+2	0	-¼	-½	-¾	0	0	0	-.01	
D31	0	0	0	0	-¼	0	-¼	-½	+.02	0	-.02	
D32	0	0	+1	0	0	0	-¼	0	0	0	0	

*Inclusion in center retarded penetration.

Material furnished courtesy Bethlehem Steel Co.

EFFECT OF PREHARDENING STRUCTURE UPON THE P-F CHARACTERISTIC

It will be noted that practically all of the tool steel samples have been given a high temperature oil quench before the final hardening operation. This has been done to remove structural ir-

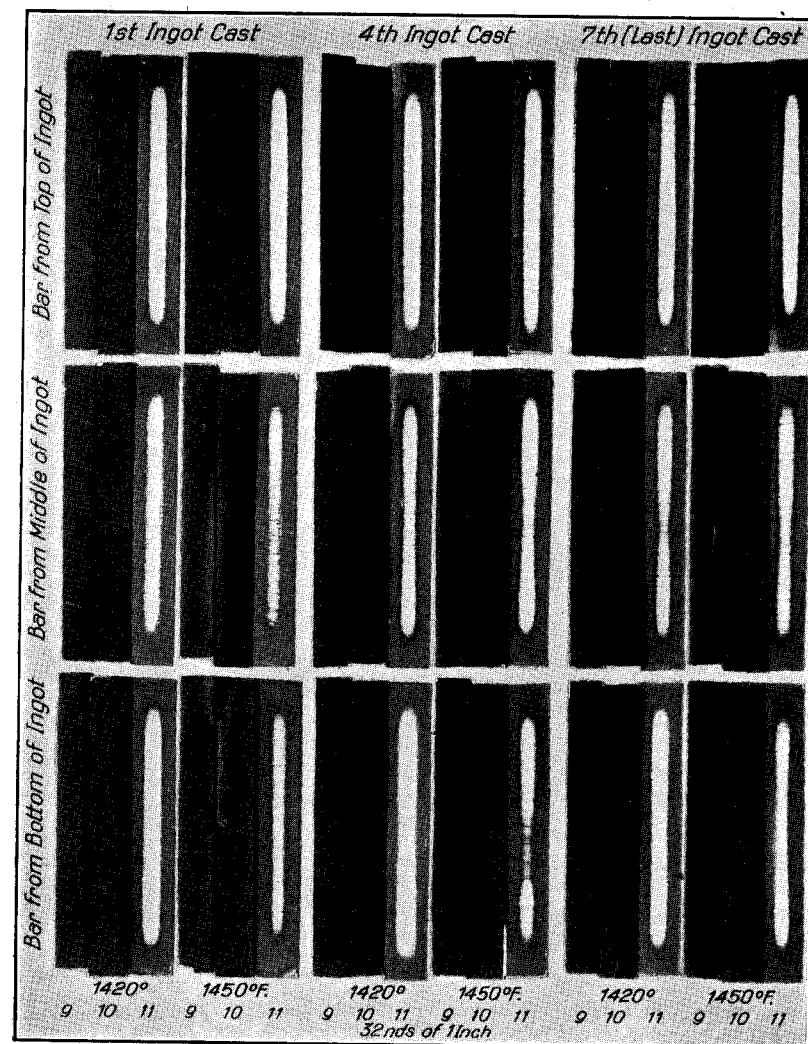


Fig. 9—Polished and Etched Specimens Showing the Uniformity of Hardenability Through the Entire Heat of Steel (Shepherd Hardenability Test). Material Courtesy Halcomb Steel Company.

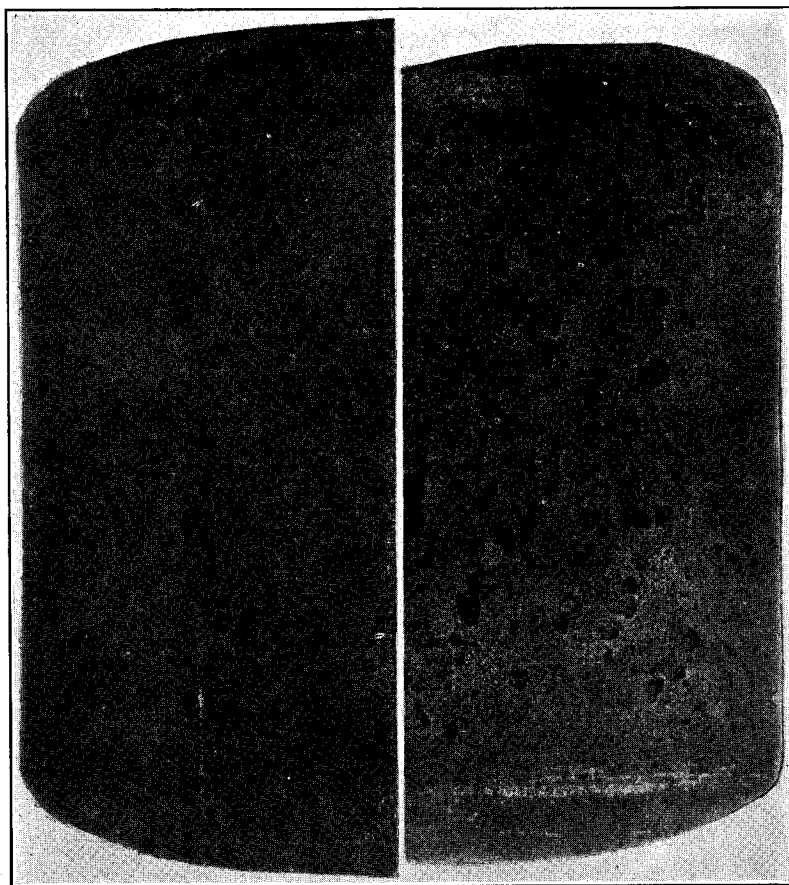


Fig. 10—Etched Sections of Half of the Cross Sections of Billets from Steel B-332. Left, Billet from Fourth Ingot. Right, Billet from Last Ingot.

regularities and was found necessary for definite control of the P-F test. Preliminary experiments showed that the P-F char-

TABLE VIII
P-F CHARACTERISTIC BETWEEN PEARLITE AND SORBITIC PREHARDENING STRUCTURES

Fig. Number	Type	Heat Number	Structure	P-F Characteristic							
				Penetration					Fracture		
15	1	30864	Sorbite	6	7	8	9	9	8¼	8	7¼
16		30864	Pearlite	7	7	7½	8	9	8¼	8¼	8
17	3	29983	Sorbite	7	8	10	24	8¼	8½	6½	4
18		29983	Pearlite	8	8	9	11	8¼	8¼	7½	7¼

acteristic of occasional heats was not sensitive to prehardening structure. The large majority of heats, however, showed a definite response to definite structure changes and this sensitivity had a definite relation to the type of steel as determined by P-F tests.

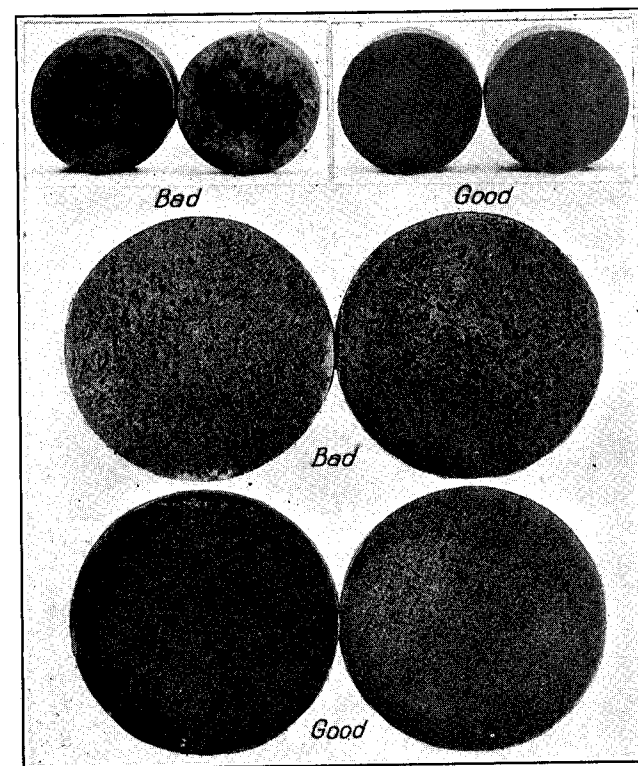


Fig. 11—Bars Rolled from Billets Shown in Fig. 10.

Consistent difference has been noted when the prehardening structure was changed from pearlite to sorbite. Heats of Type 1 characteristic were rather insensitive to changes in prehardening structure, while Type 3 heats showed a marked difference in the P-F characteristic between pearlitic and sorbitic prehardening structure. This is shown in Table VIII.

Rockwell hardness surveys show certain interesting features, see Figs. 15, 16, 17 and 18. The curves for Type 3 illustrate the points more clearly. The gradation between the martensitic and troostitic zones is much more abrupt with a sorbitic preharden-

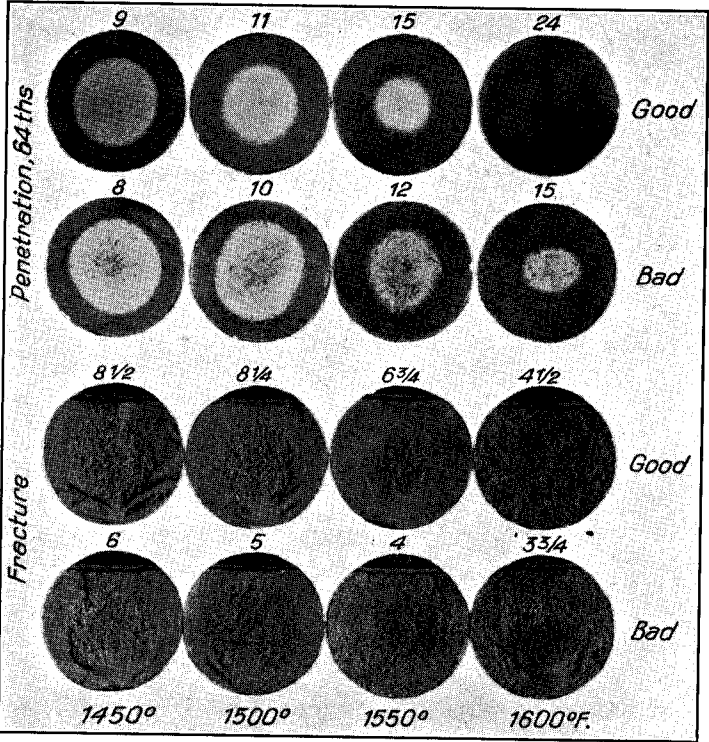


Fig. 12—Influence of Inclusions on the P-F Characteristic of Carbon Tool Steel.

TABLE IX
RELATIONSHIP BETWEEN TYPE OF STEEL AND UNIFORMITY OF HARDENED ZONE

Type	Code	Heat No.	P-F Characteristic							
			Penetration				Fracture			
1	BE	H-4974	6	7	8	9	9	9	8 1/2	8 1/2
2/1	DH	X-6903	6	7	9	15	9 1/4	8 3/4	8 1/2	8 1/4
2	CX	X-6892	5 1/2	6 1/2	8	14	8 3/4	8 1/2	8 1/4	7 1/4
2/3	M	W-5129	6 1/2	7 1/2	8	11	9 1/4	9	8 1/4	7 1/4
3	CF	H-5184	7	8	9	24	9	8 1/2	8	6 1/4
3	DN	H-5262	7	9	13	24	9	8 3/4	8 1/4	5
Chemical Composition										
Average		C	Mn	P	S	Si	Ni	Cr		
		1.06	0.23	0.013	0.014	0.20	0.08	0.04		
BE		+0.02	-0.02	0.002	0	-0.02	+0.01	0		
DH		-0.01	0	+0.003	-0.001	+0.02	-0.01	+0.01		
CX		0	0	+0.001	-0.001	+0.05	+0.01	-0.01		
M		0	-0.04	-0.001	+0.001	-0.01	-0.01	-0.01		
CF		0	+0.01	-0.002	-0.004	0	-0.01	+0.01		
DN		-0.12	+0.06	-0.001	+0.004	-0.02	0	+0.01		

ing structure than with the pearlitic prehardening structure, as the general slope of the sorbitic curves for normal hardening temperatures is steeper than the pearlitic curves. The curves also show that heating to 1600 degrees Fahr. and quenching in oil definitely reduces the penetration value over that of the same sample annealed previously at 1450 degrees Fahr. and hardened at normal temperatures. When higher temperatures (1550 and 1600 degrees Fahr.) are used, the reverse is true and the sorbitic structure tends to cause deeper penetration and coarser grain structure.

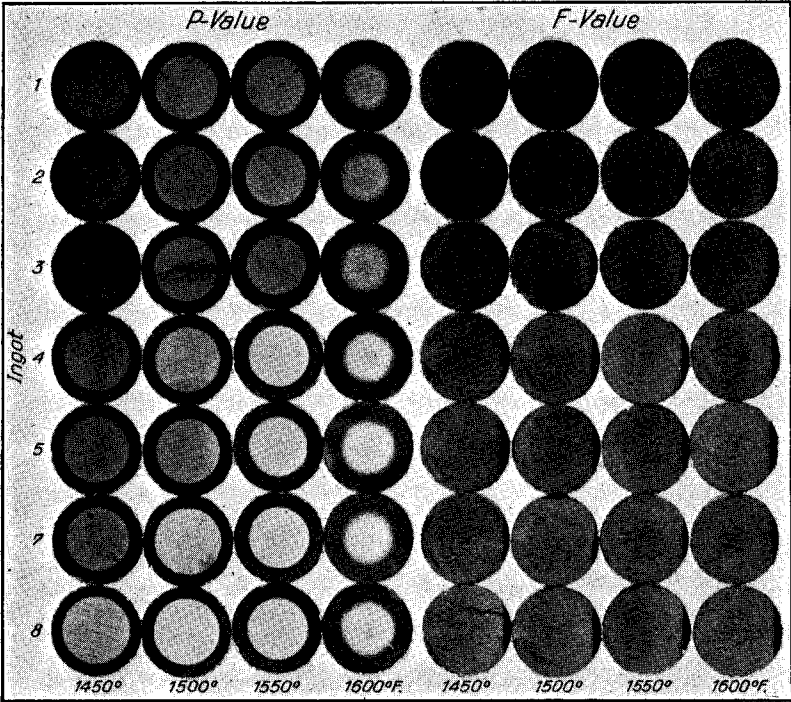


Fig. 13—Exploration of Heat 31196, Specification 1 for Uniformity of Penetration-Fracture Characteristic. (See Table VI.)

It is very difficult to obtain hardness readings close to the surface when using the cross section of P-F test pieces and a number of tests were made, using Vickers pyramid hardness testing machine. It was possible by this test to identify the first appearance of troostite, the last traces of martensite and the last traces of troostite, in the microstructure, as indicated by observation of the

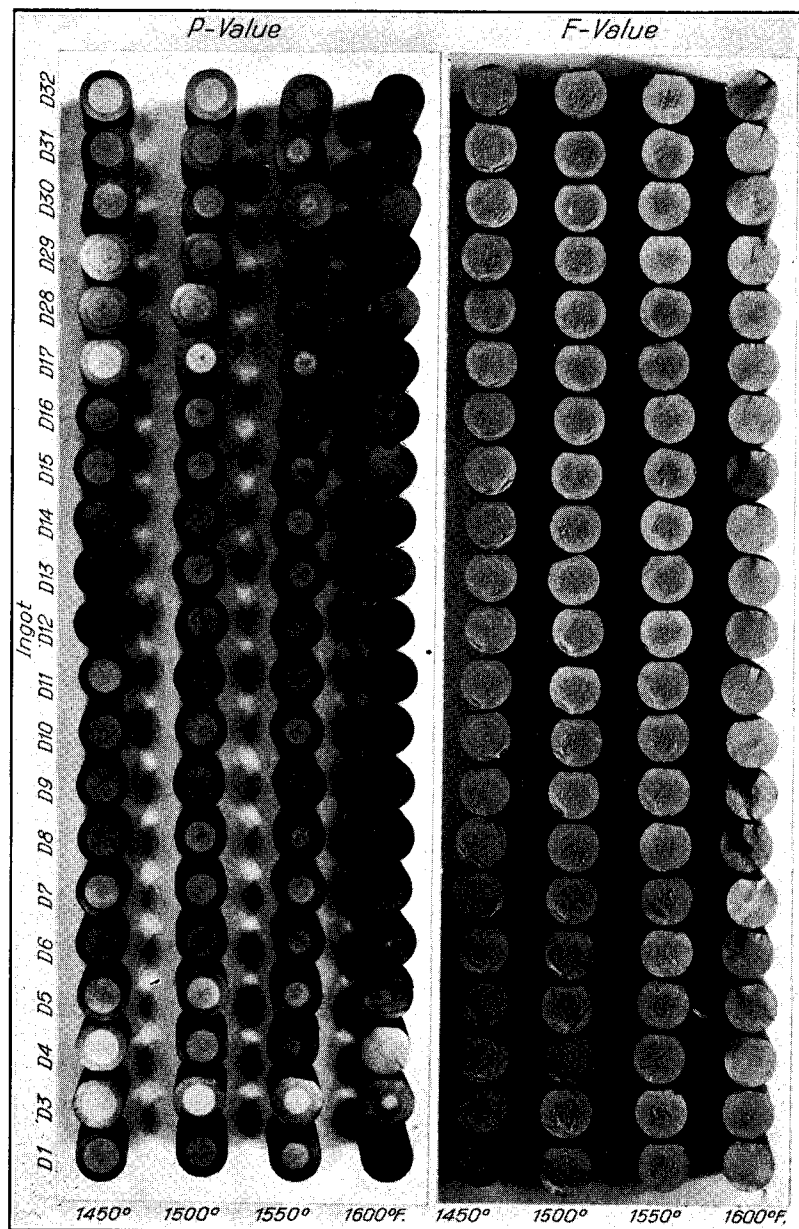


Fig. 14—Exploration of Heat 1D006; Specification 332 for Uniformity of Penetration-Fracture Characteristic. (See Table VII.)

hardness curves. The sensitivity of these tests shows a distinct tendency for Type 3 steels to have a more uniform hardness in the hardened zone than Type 1. There seems to be a definite relation

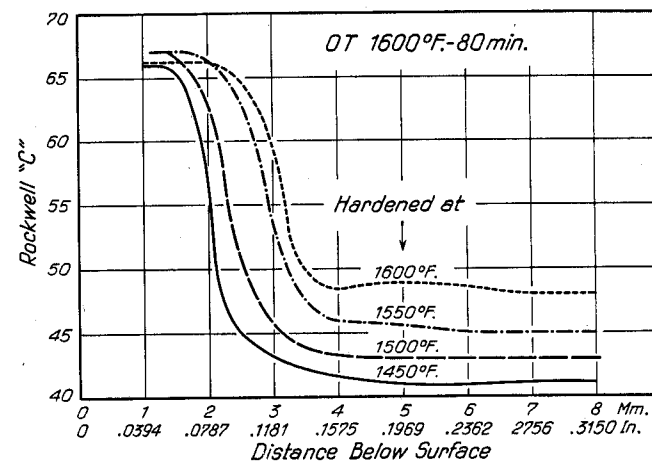


Fig. 15—Rockwell Hardness Survey of Steel X30S64 Hardened as Shown. P-F Type 1, See Table VIII.

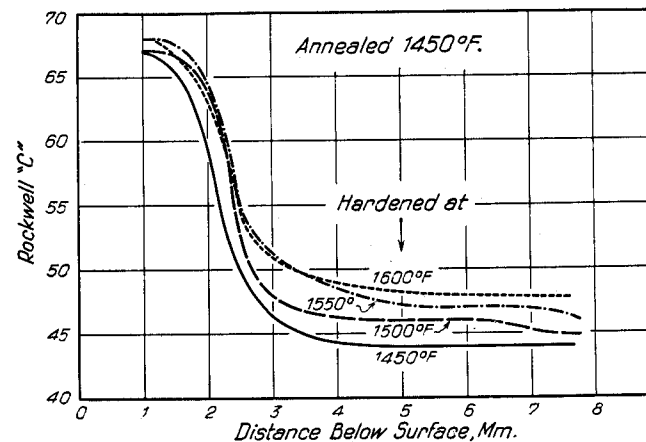


Fig. 16—Rockwell Hardness Survey of Steel X30S64 Hardened as Shown. P-F Type 1, See Table VIII.

between the type of steel and the uniformity in the hardened zone as shown in Fig. 19 and Table IX, and this feature is being subjected to further investigation.

MODIFIED P-F TEST

Grossmann* found that "each different lot of steel had its char-

*M. A. Grossman, "On Grain Size and Grain Growth," Transactions, American Society for Steel Treating, Vol. XXI, Dec. 1933, No. 12.

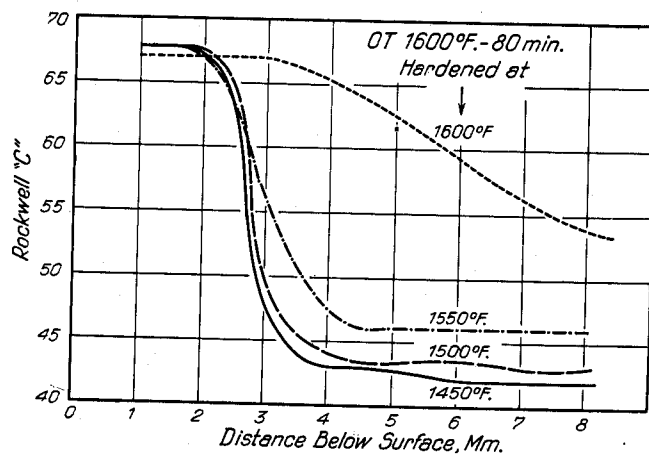


Fig. 17—Rockwell Hardness Survey of Steel X29983 Hardened as Shown. P-F Type 3, See Table VIII.

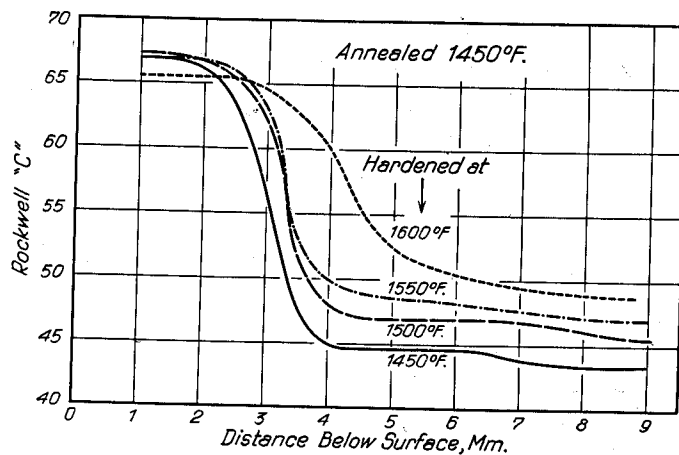


Fig. 18—Rockwell Hardness Survey of Steel X29983 Hardened as Shown. P-F Type 3, See Table VIII.

acteristic coarsening temperature.” “If the carburizing temperature is conducted at any temperature below the coarsening temperature . . . , then the austenite grain-size will remain unchanged, precisely what it was when the steel first transformed to austenite upon heating.” It is evident that the McQuaid-Ehn grain-size test requires a standard carburizing temperature for uniform results. This has been set at 1700 degrees Fahr., which is higher than usually used in commercial carburizing.

It was thought that the P-F test could be modified for carburizing and other steels so that the results would indicate the hardness, core hardness and fracture which would be obtained in commercial utilization of the steel. A similar test was made to the P-F test previously described, excepting that the samples were

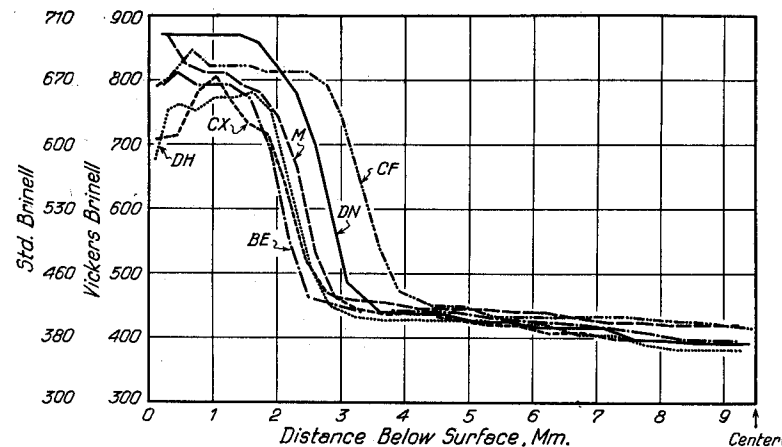


Fig. 19—Vickers Hardness Survey of $\frac{3}{4}$ -Inch Round P-F Test Pieces. See Table IX.

carburized for 16 hours after heated through at 1640 degrees Fahr. (a temperature commonly used in our factory), cooled in the pots and hardened according to the standard P-F method.

TABLE X
COMPARISON OF A.S.T.M. GRAIN-SIZE CLASSIFICATION WITH MODIFIED P-F TEST OF S.A.E. 1020 SIMPLE CARBURIZING STEEL

Manufacturer	A.S.T.M. Grain Size	Case Fracture Grain-Size Degrees Fahr.				Core Brinell Hardness Degrees Fahr.			
		1450	1500	1550	1600	1450	1500	1550	1600
C	3	6 $\frac{3}{4}$	6 $\frac{1}{4}$	6	4 $\frac{3}{4}$	192	192	207	207
A	3	6 $\frac{3}{4}$	6 $\frac{1}{4}$	5 $\frac{3}{4}$	4 $\frac{1}{2}$	207	207	229	229
A	4	7	6 $\frac{1}{2}$	6 $\frac{1}{4}$	5 $\frac{1}{2}$	212	207	235	229
A	8	7 $\frac{3}{4}$	7 $\frac{1}{2}$	7 $\frac{1}{4}$	7 $\frac{1}{4}$	207	207	229	229
B	8	7 $\frac{1}{2}$	7 $\frac{1}{4}$	7	6 $\frac{3}{4}$	201	207	235	235

Fracture grain-size tests were made of the case and Brinell hardness tests made in the center of the core.

It will be noted from Tables X and XI that the McQuaid-Ehn test bears a close relation to the fracture grain-size obtained on the modified P-F test. This indicates that observation of the frac-

TABLE XI
COMPARISON OF A.S.T.M. GRAIN-SIZE CLASSIFICATION WITH MODIFIED P-F TEST
OF S.A.E. 4615 NICKEL-MOLYBDENUM CARBURIZING STEEL

Manufacturer	A.S.T.M. Grain Size	Case Fracture Grain-Size —Degrees Fahr.—				Core Brinell Hardness —Degrees Fahr.—			
		1450	1500	1550	1600	1450	1500	1550	1600
A	.17 C	8	8¾	8¼	8¼	7¾	363	363	388
	.52 Mn								
B	.16 C	8	8¾	8¾	8¾	8¼	285	302	302
	.51 Mn								
C	.16 C	5 & 8	8¾	8	8	7	255	269	277
	.51 Mn								
E	.18 C	7	7	6¾	6	5¾	363	375	375
	.54 Mn								
D	.17 C	6	7	6¾	5¾	5	302	311	311
	.61 M n								
E	.16 C	3	6	5½	4¾	4	285	321	321
	.40 Mn								

ture grain-size of carburized articles during commercial hardening offers a quick, ready and accurate method of determining "grain-size" without reference to the microscope.

ACKNOWLEDGMENTS

Due acknowledgment is made to the metallurgical staffs of Ingersoll-Rand Co., Halcomb Steel Co., Bethlehem Steel Co., Carpenter Steel Co., and SKF Steels, Inc., for assistance in this work.

DISCUSSION

DR. JOHN JOHNSTON, *Chairman*

A. E. FOCKE, (*Research Metallurgist, Diamond Chain Co., Indianapolis:*) Three criteria which may be used to judge the merits of any test are: 1. The importance of it; 2. Its accuracy and the ease with which it may be reproduced; 3. The extent to which the purpose of it is realized.

Any test which has for its purpose an attempt to define or classify that quality in steel which is variously described as "body," "timbre," and "personality," is vitally important to a user or manufacturer of heat treated steel articles.

Mr. Shepherd has devised a relatively simple method of studying, classifying, and correlating two of the important characteristics of a heat of steel. It is obvious that he has carefully investi-

gated the limitations of his method and established a definite procedure which must be followed. However, it appears significant that the author admits that the determination of the fracture part of his P-F test by visual comparison with a set of standards requires a high degree of personal skill.

The foregoing, together with the fact that I would personally react more favorably to some test method, which would permit a grain-size analysis, without the use of an arbitrary set of standards, are my principal criticisms of the author's test method. I was somewhat disappointed that Mr. Shepherd did not illustrate the value of his test by more definite references to specific problems for which he has found it to be useful.

I am particularly concerned with the "personality" of the carburizing grades of steel and I was therefore, greatly interested in the author's modification of his P-F test to meet this requirement.

An analysis of the results recorded in Table XI reveals some rather interesting information.

There appears to be no simple relation between A.S.T.M. grain-size or for that matter Case Fracture grain-size and the core strength. Neither is there any agreement between the amounts of the common elements as indicated by the regular chemical analyses and the core hardness developed after heat treatment. Admittedly, the factors of normality and abnormality must play important parts in controlling this core reaction but measuring the relative degrees of normality of alloy steels of this class particularly in the fine grain-sizes is not simple. We have been forced to recognize these facts and have had to add core strength after hardening requirements to our specifications for certain critical parts.

It can be seen that there is much greater difference in the core strength of different steels after the same treatment than there is between the core strength of the same steel after quenching from different reheating temperatures. (Maximum difference between core Brinell of the same steel after quenching from 1450 and 1600 degrees Fahr. was 36—Minimum 17—Average 24.) (Maximum difference between core Brinell of different heats after quenching from 1600 degrees Fahr. was 111). This result confirms our experience in attempting to salvage heats which showed insufficient core strengths by simple variation in the reheating temperature.

I question, therefore, that the amount of additional information obtained from testing this type of steel at the four temperatures over that obtained from using one temperature would justify the additional labor.

But, it occurs to me to wonder if the author has not, by limiting his "penetration factor" in his modified P-F test to a simple record of the core Brinell hardness after reheating, overlooked the im-

portant variables of depth of case penetration and gradation of case hardness which, of course, are very important factors in determining the value of a part for service.

The amplification of the author's modified P-F test to include these variables appears to be to have interesting possible applications to my own problems and I therefore, wish to express my appreciation to the author. (*Written Discussion.*)

H. G. KESHIAN, (*Metallurgist, Chase Companies, Waterbury, Conn.*): The author states that hardenability is a heat characteristic. Amplifying this point further he says, "This indicates that P-F characteristic is due to the melting reactions and not to the ingot size, temperature of pouring, etc." This statement of the author would mean that the different bars of steel from a given heat will have the same hardenability characteristic. This, however, does not seem to be the case, because at times, from the same heat of steel, not only the different bars but also the different ends of the same bar may show considerable difference in hardenability, due mostly to the degree of segregation in the different ingots, and in the different ends of the same ingot. If this is so, contrary to the statement of the author, the size of the ingot, that is, the rate of solidification and the temperature of the pouring (because these influence the degree of segregation) must all have an effect on the hardenability of the material. Consequently, because a part of a heat of steel shows a satisfactory hardenability, that does not mean that the rest of the heat has the same hardening characteristic.

Furthermore, as the author very correctly states, the necessity of testing the first and last ingots from the same heat also shows that the fact that various bars of steel are made from a given ingot is no indication of uniformity, except, of course, for chemical composition. So, when considering the effect of a heat of steel from the standpoint of uniformity, some other tests such as hardenability, segregation, or macro-structure have to be applied to each bar to verify the uniformity of the various characteristics. This still leaves the steel mill practice as the best guarantee for the uniformity of hardenability, macro-structure, etc. (*Written Discussion.*)

F. H. ALLISON, JR., (*Research Metallurgist, Crucible Steel Company of America*): From the standpoint of tool steel, and particularly of carbon tool steel, Mr. Shepherd's paper is of extraordinary importance. It has long been recognized that fundamental differences exist in carbon tool steel heats melted under different practices. Mr. Metcalf, some thirty or forty years ago, on the same site at which the writer works at present, devised the Metcalf or fracture test familiar to all tool steel men, in order to distinguish between the various melts of carbon tool steel. All accredited tool steel manufacturers have long had their individual test methods to determine and classify the body or quality of tool

steel; but it has remained for Mr. Shepherd to develop and publish a method which bids fair to be accepted as a standard for the industry.

For the reason that the author's method is likely to become a standard procedure for an entire group of steel suppliers and consumers, it merits the most careful consideration and criticism in order that it may be applied as widely and fully as possible. The following remarks are offered with this purpose in view.

To be able to refer to a standard set of fractures in which a No. 9 means the same in every shop, mill and laboratory, will come as a relief to metallurgists and tool steel men who have long been troubled by the vagaries of "fine," "medium" and "coarse." Mr. Shepherd has done a service, indeed, if the graded fractures are established as a standard of reference, and the writer heartily endorses their use.

Concerning the designation of fracture, it is felt that the term "grain-size" should be avoided, since this term already has a significance in metallography which easily leads to confusion if adopted also in reference to fracture. The term "fracture grain-size" is little better since in discussion it is extremely easy to omit the word "fracture." It is suggested that "fracture appearance," "fracture texture" or some better term be used for the sake of clarity in these references.

In regard to penetration values, the author reports them in terms of $\frac{1}{64}$ th inch or half of $\frac{1}{64}$ th inch. There are advantages in reporting these measurements in $\frac{1}{100}$ th inch and these should be carefully considered before the method is too rigidly standardized. The accuracy of measurement is about the same in using either $\frac{1}{128}$ th inch or $\frac{1}{100}$ th inch. By reporting the true hundredths of an inch penetration the use of fractions is avoided, the magnitude is apparent in the actual figures, and the decimals readily distinguish the penetration from the fracture values in the full P-F report. Thus a steel reported by the present system as:

5½, 6½, 8, 14 : 8¾, 8½, 8¼, 7¼

would become:

.09, .10, .13, .22 : 8¾, 8½, 8¼, 7¼.

The gage shown in Fig. 4 of the author's paper could be as easily calibrated in hundreds as in sixty-fourths of an inch.

Further work in regard to penetration measurement appears to be necessary in that the transition zone between the hardened case and core may, as the author points out, be several hundredths of an inch in width, so that the exact boundary of the hardened zone is not always easily determined.

The splendid conception of grouping the change in penetration with the change in fracture appearance as the hardening temper-

ature is increased, affords a very full picture of the behavior of the steel. The further classification of steels into types such as 1, 2/1, 3, etc., may, however, be satisfactory for individual application and for simplification; but, for general application it appears with the present knowledge to be somewhat premature and may also be misleading. In the first place, it is felt that wider experience over a greater number of tool steel applications should be obtained before fixing the arbitrary limiting values assigned to the several types. Secondly, a single type as it now stands may embrace steels of widely different character. For example, the P-F characteristics of the following two steels would both be classified as type 1, yet the first is shallow hardening with fine fracture, and the second is relatively deep hardening with coarse fracture.

	Penetration	Fracture
First	.08, .08, .10, .11	: 9, 9, 9, 8½
Second	.14, .14, .16, .17	: 6, 6, 6, 6½

The author calls attention to the dumb-bell type of core in Fig. 8 found on hardening certain disks. Fig. 9 indicates that this condition is more likely to be characteristic of the portion of the ingot from which the sample was taken, rather than characteristic of the entire heat.

The paper contains data which are decidedly pertinent to the general subject of hardenability. The supposition has been advanced by others that the depth of penetration depends upon austenitic grain-size and that the greater the grain-size, the deeper will be the hardened case. Similarly, it has been assumed that the fracture appearance is due to the austenitic grain-size—the coarser fracture being occasioned by larger original austenitic grains. However, from the data of the present paper, it is possible to find a shallow hardening steel possessing a coarser fracture than a deep hardening steel or vice versa. In fact, the depth of hardening and the fracture appearance seem to be entirely independent of one another as can be seen from the following data selected from the paper:

Heat	Penetration	Fracture
H5194	.09, .09, .11, .13	: 9, 9, 9, 8½
5760	.08, .09, .10, .12	: 8½, 7¾, 5, 4½
2 H	.22, .22, .22, .23	: 8, 8, 7¾, 7¾
30420	.14, .17, .38, .38	: 8¾, 7¾, 6¾, 5½

Evidently, the growth of the austenite grains cannot be the cause of the difference both in hardenability and in fracture, or there would be more correlation between the two than above indicated.

In concluding, the author has made a valuable contribution to the metallurgy of tool steel, both in providing an excellent test

method and in emphasizing that there exist a number of types of carbon tool steel, independent of composition, that are available to the tool steel user to suit his various requirements. (*Written Discussion.*)

R. SCHEMPP, (*Metallurgist, Halcomb Steel Co., Syracuse, N. Y.*): In his present publication, Mr. Shepherd has given both the steel maker and the consumer another valuable and very practical means for the determination of the inherent hardening, grain-size and temperature sensitivity characteristics of steels, especially tool steels. Unlike the rather unfortunate choosing of "normal" and "abnormal" as a means of conveying the structural grain-size characteristics of the hypereutectoid zone as developed after the standard McQuaid-Ehn carburizing treatment and observed under the microscope, Mr. Shepherd's terminology and method of designation as well as his test procedure, distinguish themselves by their extreme clearness, simplicity and accuracy and lend themselves to a quantitative measure of the properties involved.

Speaking of the inherent hardenability and P-F characteristics of tool steels of identical composition, it may be well to point out that while these properties are essentially the result of certain melting practices, these so-called inherent characteristics of hardenability and martensitic penetration upon quenching, with but few exceptions (that is, on very stable types) are materially influenced by

1. Prior structure.
2. Critical rate of cooling during the quenching operation.
3. Mass action.

The apparent contradictory influence of prior structure upon the hardening properties of tool steels when using a low (1450 degrees Fahr. or high 1600 degrees Fahr.) hardening temperature, may be explained by the predominating influence of carbide solubility as brought about by prior structure in one case and by the stronger influence of the austenitic grain-size prior to quenching in the other.

The early results of test work which is being carried on in our laboratory indicates a very definite relationship between the austenite grain characteristics and the hardenability and P-F values as determined by the Shepherd methods. On tool steels, Mr. Shepherd's testing procedure will allow us to definitely classify heats as to their specific types. Our practice has been to differentiate between the following four types:

1. Material on which the penetration of hardened case as well as the grain-size is stable, and will not increase with increasing temperature.

2. Material on which the penetration of hardened case will increase in depth with increasing temperature, the grain-size of which, however, will not coarsen.
3. Material which will not increase in penetration of hardened case with increased quenching temperature, but the grain-size of which will coarsen.
4. Material on which both the penetration of hardened case and grain-size will increase with increasing quenching temperature.

Neither by this means nor by any other means of testing which has been developed in the past has it been possible to determine the physical property characteristics as ductility, toughness, etc., of this class of material in the hardened condition in such a manner as would enable us to differentiate between heats of similar composition in regard to their toughness in the hardened state, and to predict certain service results for various applications.

In his efforts to develop a more satisfactory heat treatment for our high speed steel grades, Mr. Kingsbury, a member of our metallurgical staff, applied as early as 1911 the static load fiber stress test as later described by E. G. Herbert under "The Influence of Heat on Hardened Tool Steels" in the *Journal of the Iron and Steel Institute*, Vol. 85, #1, of 1912, p. 358 to 378, to determine the toughness characteristics of high speed steel. Herbert's determinations were primarily confined to maximum fiber strength alone. Strength, however, is not necessarily a measure of toughness. This comparatively simple means of obtaining the maximum fiber stress values on hardened material has now been modified and developed into a testing procedure which lends itself remarkably well for an actual determination of the inherent toughness values of hardened high speed and tool steels.

This essential "it" of tool steels, may it be termed toughness, body or any other name, is an inherent heat characteristic which all of us have learned to know and to value, but for which we have never found a means of expression and determination. Our new method tests the outermost fibers of the specimen under conditions of comparatively heavy and rapid loading and while our test is admittedly a static test as against the well recognized fact that tools in service fail through dynamic stresses, we believe that our method very closely approximates actual service conditions.

In our opinion toughness must be considered as a combination of strength and ductility and, as we will show in the following results which have been obtained on several heats of straight carbon, general purpose tool steel, we have expressed this inherent toughness in the form of a coefficient of toughness value.

To eliminate the influence of prior structure, we also resort, as does Mr. Shepherd, to a preliminary oil treatment before harden-

ing our specimens which are usually $\frac{3}{8}$ -inch round and approximately 6 inches long. In the case cited the samples were hardened from 1550 degrees Fahr. and tempered at 415 degrees Fahr. for one-half hour.

In the following table we give the composition of the material used, its characteristics as obtained by the Shepherd hardenability and P-F method as well as the normality and grain-size rating determined according to the McQuaid-Ehn test.

Composition:

Heat No.	C	Mn	P	S	Si
H-4705	1.08	0.23	0.017	0.018	0.29
W-5294	1.05	0.21	0.011	0.015	0.24

Results of regular testing procedure:

Heat No.	Hardenability Rating (slab test)	P Value	F Value	Type	McQuaid-Ehn Test (A.S.T.M. Chart)
H-4705	10	6, 7, 8, 12	8 $\frac{3}{4}$, 8 $\frac{1}{2}$, 8 $\frac{1}{4}$	2/1	Abnormal 6/7 Grain
W-5294	10	7, 8, 10, 24	8 $\frac{3}{4}$, 8 $\frac{1}{4}$, 7, 6 $\frac{1}{4}$	3	Abnormal 5/6 Grain

Kingsbury Toughness Test:

Heat	Fiber Strength	Ultimate Deflection	Coefficient of Toughness	Rockwell Hardness
H-4705	222,500	0.094	21,150	62.5
W-5294	181,000	0.0674	12,200	62

We may add that Heat H-4705 was reported as being very satisfactory and tough bodied in service; while the material of Heat W-5294 proved brittle and unsatisfactory.

This new test has given us a number of very convincing confirmations from its practical application. We have mentioned this new test because it seems to us that it supplements Mr. Shepherd's hardenability and P-F testing procedure.

Complete details of this toughness test for hardened tool steel materials will be published in the near future. (*Written Discussion.*)

L. D. HAWKRIDGE, (*Hawkridge Brothers Company, Boston, Mass.*): Again we are indebted to Mr. Shepherd for a most interesting and important contribution, disclosing a simple and thoroughly practical method of classifying tool steel in accordance with the hardening characteristics—a method, which, in the writer's opinion, will take its place beside the McQuaid-Ehn test as one of the indispensable tools of the industry.

While the manufacturer or distributor of tool steel may deplore the advent of another test containing a large measure of the "personal equation" (which there undoubtedly is in the estimating of the F values), he must admit that by no other means known at present could so much, or so exact information, be packed into so little space as by the series of eight numerals proposed by Mr. Shepherd.

The latter himself wisely stresses the difficulty of rating fractures correctly, and advocates special training of those who will act as observers. This is important, as the writer can testify from the few attempts that he has made to score over 80 per cent in the fracture rating test. Mr. Shepherd might well have stressed also the fact that luster often tends to obscure grain-size. It is a common error of beginners to rate a fiery fracture too far toward the coarse end of the scale. But in spite of the difficulties that one can foresee in the application of this test, it will be by no means more difficult to obtain agreement between buyer and seller than in other tests now widely used—for instance, the hot acid etch test.

Of particular interest to the writer is the comparison between the P-F values of the 8-ton basic electric heat and the 100-ton basic open-hearth heat, as given in Tables VI and VII. How does Mr. Shepherd account for the apparent instability both of the grain-size and penetration of the specimens from the open-hearth heat which were hardened from 1550 degrees Fahr.? Here is a variation of 50 per cent in the penetration of hardness in specimen D 7 as against D 4, whereas the maximum variation between ingots of the electric heat at the same temperature is less than 10 per cent. Why should the grain-size vary so widely between specimens of the open-hearth heat quenched from 1550 degrees Fahr. when those quenched from 1600 degrees Fahr. are in much closer accord?

One more question—Do not the results set forth in Table VIII indicate that, when a test is being made of Type 3 steel, it is most important to maintain the specimens at 1600 in the normalizing for the full 80 minutes recommended earlier in the paper? If this soaking time is cut down, will not results be questionable because of non-homogeneity of the pre-hardening structure? Incidentally, why should the relatively coarse particles of carbide in pearlite (presumably spheroidized) inhibit grain growth at high temperatures more than the very finely divided carbides in sorbite?

While particularly applicable to tool steel, Mr. Shepherd's method, as modified for carburizing steels, will be decidedly useful, and there would seem to be no reason why a similar modification could not be worked out for steel of intermediate carbon ranges both with and without alloys. With such broad fields for further investigation opening out, we should be able to look forward to many more illuminating papers from Mr. Shepherd on the same general theme. (*Written Discussion.*)

G. V. LUERSSEN AND O. V. GREENE, (*Metallurgy Department, The Carpenter Steel Co., Reading, Pa.*): Probably no physical characteristic of steel is more actively discussed and less accurately reported than the fracture test. We are certainly indebted to Mr.

Shepherd for giving us standards for the evaluation of a very useful test.

The phenomenon observed by Mr. Shepherd on specimens pre-hardened by oil treating from 1600 degrees Fahr. is particularly interesting. Table VIII, which reports P-F characteristics on types 1 and 3, shows that heating specimens to 1600 degrees Fahr. and oil quenching, followed by brine quenching, produced a shallower penetration than was obtained in specimens brine-quenched from the annealed condition. This was true only at temperatures of 1450 and 1500 degrees Fahr. After this temperature range was passed, and higher temperatures such as 1550 and 1600 degrees Fahr. were employed, the opposite was the case, and the preliminary oil treatment tended to cause a deeper penetration and a coarser fracture. These characteristics were particularly noticeable in the heat of type 3.

It is quite possible that these results may be explained by considering the structure of these specimens before the final hardening. A steel of type 1 after oil treating probably retains some of the eutectoid carbide in solution, with the balance submicroscopic or nearly so in size. This treatment also disperses and reduces the size of the hypereutectoid carbide. Consequently, when a specimen of type 1, oil treated from 1600 degrees Fahr., is heated for final brine quenching from 1450 degrees Fahr., for instance, the first austenite will form where the greatest amount of carbides are in solution or where the spheroids are finest. These early crystals of austenite must, therefore, be exceedingly small and very numerous. The balance of the ferrite and some of the undissolved carbide will then be absorbed as the austenite grows. However, at this temperature there will remain a comparatively large number of widely distributed carbides. These well dispersed particles not only impede the grain growth, but act as nuclei which accelerate the transformation and, consequently, produce a decrease in the hardenability.

When an annealed specimen of a steel of type 1 is heated for hardening at 1450 degrees Fahr., a similar reaction takes place, except that the formation of the austenite is considerably more difficult, and the original islands are fewer. The undissolved carbides impede the grain growth as before, but are not sufficiently dispersed to offer enough nuclei to accelerate the transformation. Consequently a slightly deeper hardenability is exhibited.

As a matter of fact the rate of transformation of oil-treated specimens on heating is so increased, or the actual temperature of transformation lowered, as reported by Mr. Shepherd, that it can be easily observed in an ordinary heating furnace. If an oil-treated and an annealed specimen be placed in a furnace at 1450 degrees Fahr., and observations made during the heating, it will be noticed that the oil-treated specimen will be through its critical and nearly up to furnace temperature, while the annealed specimen will still appear quite black.

The structural changes occurring in an oil-treated specimen of type 1 steel during heating for final brine quenching at 1600 degrees Fahr. are similar to those that took place when an oil-treated specimen was heated to 1450 degrees Fahr. However, as the larger austenite islands grow, they absorb more and more of the smaller and less favorably situated islands, until all the grains have reached a fairly uniform size. The eutectoid and hypereutectoid carbides in this specimen are so small and so well dispersed that they are readily and completely soluble at 1600 degrees Fahr. Consequently with no particles to arrest the grain growth or act as points of crystallization on quenching, the specimen exhibits a deeper hardenability and a coarser fracture than the one brine-quenched from 1450 degrees Fahr. in the oil-treated condition.

The annealed specimen of type 1, on the other hand, when heated for brine-quenching at 1600 degrees Fahr. does not dissolve all of the hypereutectoid carbide. The hypereutectoid carbides in this specimen are comparatively large and difficultly soluble. They impede the grain growth on heating and offer points of crystallization on quenching. Consequently the presence of the carbides in the annealed specimen brine-quenched from 1600 degrees Fahr. produces a finer grain and a shallower hardenability than can be obtained in the oil-treated specimen brine-quenched from the same temperature.

The dissimilarity between the P-F characteristics of the oil-treated and the annealed specimens of a steel of type 3 is due to the same causes and are differences of degree and not of kind.

Type 3 steels exhibit a much deeper hardness penetration and a much coarser fracture than type 1 steels, because of inherent differences. The repression of grain growth in type 1 steel, particularly in the high temperature range, is due to the presence of widely dispersed submicroscopic nonmetallic particles. Such particles do not dissolve or coalesce at the higher treating temperatures as do carbides, and, consequently, they continue to obstruct the grain growth. These particles also offer nuclei for the transformation of austenite on quenching. Thus these minute nonmetallics impede the grain growth and accelerate the austenite transformation to such an extent that a comparatively fine fracture and shallow penetration is obtained when a steel of type 1 is quenched from the higher range. A steel of type 3, which is dependent upon the presence of carbides alone for the impedance to grain growth and for the acceleration of the austenite transformation, shows a coarser fracture and a deeper hardness penetration when quenched from the higher treating temperatures. (*Written Discussion.*)

A. H. D'ARCAMBAL, (*Consulting Metallurgist, Pratt and Whitney Company, Hartford, Conn.*): Our knowledge of tool steels has been greatly increased during the past few years as a result of the

published research work of such able investigators as Shepherd, Gill, Emmons, Luerksen, Wills and others. Mr. Shepherd's paper on "The P-F Characteristic of Steel," just presented, should prove of interest and value both to manufacturers and users of tool steels. He again definitely shows that steels of the same chemical composition when subjected to similar hardening treatments react differently, certain steels hardening much deeper than others, a coarser grained fracture also being noted on some of the types tested. We would be interested in learning the composition and the treatment given each of the ten specimens shown in Fig. 1. Another set of masters made from a different heat of steel of the same composition, especially if produced by a different mill, undoubtedly would be quite different in fracture grain-size for corresponding samples. Is it not quite difficult, moreover, for two observers to check the "F" factor within half a fracture number?

We would be interested in knowing if all of the carbon tool steel purchased by Mr. Shepherd's company is subjected to this so-called "P-F" test. Should this be the case, has not the author found that tool steel of the same chemical composition from the same mill very rarely does not pass this "P-F" test from heat to heat. We have found that tool steel, heat after heat, from our principal source of supply shows a satisfactory fracture and hardness depth when quenched from elevated temperatures, but tool steel of the desired composition from other mills will not always react so favorably to these valuable tests first suggested by the author some years ago.

We noted with considerable interest Mr. Shepherd's comments relative to a comparison between electric furnace and lead pot method of heating. It has been our experience that samples heated in the lead bath come to temperature very rapidly as compared with similar sized parts heated in the electric oven type furnace. For example, tool steel specimens $\frac{3}{4}$ inch diameter by 6 inches long will harden satisfactorily after being heated in the lead bath at 1460 degrees Fahr. for only one minute. Similar sized specimens from the same bar of steel when heated in the electric furnace required 15 minutes to reach the furnace temperature of 1450 degrees Fahr. The specimens heated in the lead bath as well as the electric oven furnace treated samples were brine-quenched, after the one-minute and fifteen-minute times, respectively, resulting in a Rockwell hardness in each case of C-66-67. The samples treated in the electric furnace showed a slightly greater hardening depth than the samples given one minute in the electrically heated lead bath furnace. Specimens given longer periods of time in the lead bath showed the same hardening depth as the specimens heated for one minute.

It will be interesting to note the reaction of the tool steel manufacturers and users to the author's suggestion relative to purchasing tool steels on the penetration-fracture basis as outlined

in this interesting paper by Mr. Shepherd. (*Written Discussion.*)

S. C. SPALDING, (*Metallurgist, American Brass Co., Waterbury, Conn.*): The author is to be congratulated on originating a practical method of classification of carbon tool steels. Although tool steel makers and users have talked of inherent quality, body, timbre, etc., it was apparently left to Mr. Shepherd to give us a practical and simple method and set of standards for their determination. The general adoption of this method of classification should help materially to bring up the average quality of our carbon tool steels. A consumer today finds only to his sorrow that steels of the same analysis and supposed quality purchased from different sources vary widely in physical characteristics. When he and the mill metallurgist get together there is seldom an agreement, even in terms to describe the fracture characteristics.

Here we have a set of standards set up for us. With one of these boxes at hand the matters of difference could readily be settled. We hope the steel producers will be sufficiently impressed by this paper to equip themselves with the standard fracture box and grade their steels accordingly.

It would seem to us a great step forward to be able to purchase steel to a certain range of P-F characteristics, rather than to a brand name. I would like to recommend that our society of itself, or in conjunction with the A.S.T.M., make a study of this method, with the idea of working out acceptable standards for inclusion in specifications for tool steels. (*Written Discussion.*)

A. C. JONES, (*Research Engineer, Lebanon Steel Foundry, Lebanon, Pa.*): Mr. Shepherd's contribution to that phase of metallurgical science dealing with grain-size control and hardenability is not only interesting but extremely valuable. His whole scheme, cleverly worked out, must appeal not only to the research engineer, but also to the control metallurgist, and anyone interested in getting the most out of his steel product.

To get the evidence that Mr. Shepherd has presented, was probably less difficult than to explain the causes of the uniformities shown. Undoubtedly the method and degree of deoxidation and degasification in the melting history of the steel are very important factors. His scheme applied to the observations in the steel plant should help establish the important factors affecting the grain-size and hardenability.

I would like to ask Mr. Shepherd if any relation has been observed by him between "inclusion count" and grain-size or hardenability. Also, was there any notable difference in effect between steels with large inclusions and those with small inclusions? Another question—has he any data to indicate that the fine-grained carbon steel preserves most of its impact resistance at sub-zero temperatures? (*Written Discussion.*)

HENRY WYSOR, (*Metallurgical Engineer, Bethlehem Steel Company, Bethlehem, Pa.*): We are greatly indebted to the author who has given us a new definition for steel quality, based on obvious phenomena of hardening. The experimental work by which the data was obtained is admirable in both the extent and the care to which it was prosecuted. While the primary object was to test steel quality in the improvement of heavy duty tools, the usefulness of the "P-F" test in other lines of developments and research is foreshadowed.

It should prove very interesting to compare the "P-F" characteristics of carbon and special steels with fatigue values. Take for example spring steels, especially carbon and low alloy spring steels. There appears to be a decided connection between penetration characteristics in quenching and endurance of springs in vibratory tests provided the section is heavy enough to develop a core fracture. That is simply an observation which I have made on both spiral and leaf type springs, but I propose to apply Mr. Shepherd's method of testing for a more definite proof. (*Written Discussion.*)

THOMAS G. FOULKES, (*Metallurgical Supervisor, Bethlehem Steel Company, Bethlehem, Pa.*): The test, so fully developed by Mr. Shepherd for determining the so-called P-F characteristic of steel, and so clearly presented in his paper on this subject, will have an especial appeal to the heat treatment shop man. Here is a test that he can understand and appreciate. This test eliminates, for him, the mystery of a laboratory report couched in scientific terms, as it gives him an instrument for visual classification.

The testing procedure, while exacting, is not difficult. The adoption of this test will give impetus to improvements in the application of steel for treated parts, making possible the consistent use of a selected type of steel based on P-F values, determined from comparison of parts having given satisfactory as well as unsatisfactory service.

It is realized that all types will be in demand, the type desired depending on the particular application involved. The steel industry stands ready to co-operate in the development of the proper type for particular applications.

The writer is acquainted with the consistent production of two of these types, listed by Mr. Shepherd, over a period of several years.

The Shepherd P-F characteristic test is a distinct contribution to the advancement of steel industry technique, giving a ready and proved method for classification and selection of steels. (*Written Discussion.*)

MR. DOAN: Is there any correlation between the P-F characteristic and the gas content of the steel in question? For example, is

there any relationship between the oxygen and nitrogen content of the steel and its P-F characteristic? Furthermore, have you found that there is any relationship between the above mentioned characteristic and the grain-size of steel as measured in any of the conventional ways?

DR. J. A. MATHEWS, (*Vice-President, Crucible Steel Company of America, New York City*): In reference to Fig. 2, where Mr. Shepherd refers to normalizing as samples heated at 1600 degrees Fahr. and quenched, I think that should be called pre-quenched. The society of which Mr. Shepherd is now president, and two other societies, have a committee on definitions, but I think we should stick by our old definitions.

MR. SHEPHERD: In reply to Mr. Focke's discussion may we say that our particular applications of the P-F characteristic test are in a field which is very highly specialized and would not be of direct value to any one except those few persons engaged in that particular industry.

The applications of the test, must, therefore, be worked out for each individual's particular requirements.

The use of four temperatures serves to accentuate any differences which may exist and the test in reality grew from the inability of one temperature to give sufficient information. We believe that the test has not, as Mr. Focke states, been developed beyond its initial stages. The width of its application can only be determined by further work.

As Mr. Keshian indicates it would be quite easy to find differences in hardenability and P-F characteristic throughout a heat. We must, however, consider that the particular heats under observation represent carefully refined, melted, poured and rolled steel in which good practice has been used to such an extent as to control all of these factors within the limits which are possible at the present time. The acceptable bars from heats of steel of this type would show remarkable uniformity with regard to hardenability and P-F characteristic. If any marked non-uniformity would be shown by these tests it would indicate that a change in the source of supply should deserve careful consideration.

As Mr. Keshian states, the steel mill practice is the best guarantee of the uniformity of hardenability, microstructure, etc., but it is very necessary to continually check the results of this practice by incoming material inspection tests.

Mr. Keshian has touched on one point which is probably not emphasized or explained in detail in the original paper. This new test method is primarily used to determine the difference between different heats of steel of approximately the same chemical composition with the idea of finally correlating service results with a particular P-F characteristic. It is essential, therefore, to use

metal representative of the heat of material to be tested and for this reason we forge the stock to approximately $\frac{7}{8}$ -inch round. Tests made on an ingot from the outside and interior of an ingot, from the outside and interior of a billet, from the outside and interior of a bar, would very probably show a considerable difference in P-F characteristic, diminishing as the steel was reduced in size. For this reason any tests made in a manner different than that described should be accompanied by specific detailed data.

In reply to Mr. Allison's discussion may we say that fracture grain-size is used to designate the texture of the fracture of the P-F test pieces because it is closely allied to the actual martensitic grain-size of the test piece. There is no more excuse for omitting the word fracture in referring to the fracture grain-size, than there is for omitting the word austenitic in speaking of austenitic grain-size, or in omitting A.S.T.M. in referring to A.S.T.M. grain-size classifications, etc. It does not appeal to us to refer to the fractures as fracture appearance #8 or fracture test #6. The grain-size is an actual grain-size, although the scale is empirical.

The penetration values are reported in $\frac{1}{64}$ of an inch because the original Shepherd slab test known to the steel trade for several years was based on $\frac{1}{32}$ -inch variations in thickness. Since the penetration is measured on one side only, it was only natural that we should continue our designation of hardenability in the same terms in which it was originally conceived, namely $\frac{1}{2}$ of $\frac{1}{32}$ or $\frac{1}{64}$.

The suggestion of one hundredths of an inch should be an optional choice with the user of the test. We have accepted Mr. Allison's suggestion and are grading our gage in both scales.

It is true that single type may embrace steel of widely different analysis but the P-F characteristic type may be the same and it is felt that the classification of the P-F values into types is justified through its clear indication as to whether the steel is stable or unstable with regard to either penetration or fracture grain-size, or both. For example, we may have a 0.70 per cent carbon tool steel which is the same type as a 1.05 carbon tool steel, but its P-F characteristic will be different. It has been found that certain of the inherent properties of these steels are similar in regard to toughness, etc., but, of course, they may be of entirely different hardenability and P-F characteristic throughout.

In reply to Mr. R. Schempp's discussion we look forward with interest to complete details of the toughness test for hardened tool steel material, particularly as these materials will be identified by their particular P-F characteristic.

After this information has been published it is entirely probable that a great many of our tool steel users will see much more clearly the necessity for obtaining data on tool behavior of steels which have been identified by their P-F characteristic and not by

the indefinite approximate chemical composition. After such correlation it is entirely reasonable to expect that new specifications for tool steel will have as their essential requirement a test similar to that described in the paper.

In reply to Mr. Hawkrige's discussion may we say that unless the observers have had experience in judging tool steel fractures it, of course, is too much to expect to have a novice read fractures with the skill required by the fracture portion of this test. It has been necessary to apply special training of probably a much more intensive character to our students in metallography. Skilled tool steel men usually have very little trouble in passing 80 per cent in the fracture rating test, provided proper light conditions have been available.

We believe luster is the result of grain-size. The shininess of a surface depends upon its ability to reflect light in unbroken wave fronts. Fine-grained specimens have a tendency to be more lustrous than coarse-grained specimens.

The specimens from the open-hearth heat shown in Tables VI and VII were hardened through. 1550 degrees Fahr. was a border line where a more or less degree of variation in hardness might be expected. The electric furnace heat shown in Fig. 13 is of a different type and hardens much less deeply at 1550 and 1600 degrees Fahr. The less the depth of hardening the more nearly the specimens may be expected to be in accord, all other factors being equal. We have found it necessary to maintain the specimens at the 1600 degrees Fahr. normalizing temperature for at least 40 minutes after heated through. We agree with Mr. Hawkrige that a modification of the P-F test would be extremely valuable for steel of intermediate carbon ranges, both with and without alloys. We have offered the test with the hopes that some of our more able investigators will carry this work on.

In reply to Messrs. Luerssen and Greene's discussion there is a question in our mind whether Type 1 material produces a less soluble carbide or whether the solute is less active. Regardless of this, we are convinced that Type 1 material is not "carbide sensitive" because the presence of other penetration propagation inhibitors, such as submicroscopic oxides, etc., is more pronounced than in the other P-F types of steel of the same composition. Type 1 is not annealing sensitive and, therefore, the question of carbide size does not enter the picture.

The balance of Mr. Luerssen and Mr. Greene's discussion serves to answer the query of Mr. Hawkrige.

In reply to Mr. A. H. d'Arcambal's discussion the set of Fracture Standards is an empirical set of standards and cannot be produced by a definite material heat treatment formula. While this formula is essential as a guide, the fractures obtained must be discarded if they do not come up to the standard of accept-

ability established for each particular number. Slight variations are made in treatment during their production and a great many discards must be made for each acceptable specimen.

Another mill could use entirely different steels and heat treatments and would accomplish the same final results if they were working to the same set of primary standards. We feel that it would be very desirable for the steel fraternity to concentrate on this set of standards, which idea was also expressed in Mr. Spalding's discussion.

It is not difficult for two observers to check the fracture factor within $\frac{1}{2}$ fracture number. We ordinarily accept $\frac{1}{4}$ of a number, plus or minus, as a check, but it is not unusual for trained observers to continually check fractures to the identical reading.

All of the tool steel purchased by the Ingersoll-Rand Co. is subjected to this P-F test. We are obtaining a correlation between service performance and the P-F characteristics which will, at a later date, be transformed into a specification for P-F characteristics for the various tools.

The curves given in our heating tests represent the temperature on the inside of a piece of $\frac{3}{4}$ -inch round. It is readily conceivable that the outside of a piece of $\frac{3}{4}$ -inch round will heat in the short time mentioned by Mr. d'Arcambal sufficiently above the critical temperature to obtain a definite hardened case. This differential in temperature uniformity throughout the piece would also result in a faster cooling rate due to the non-necessity for the cooling medium to extra heat from the "cold" inside.

We have also definitely noted the difference in depth of hardness obtained as a function of the length of time through the critical temperature and at the quenching temperature. This resulted in the establishment of rigid hardening procedure for the conduction of the P-F test.

In reply to the discussion by Mr. A. C. Jones there is a noticeable difference in the hardening effect obtained on steel with large inclusions, as will be noticed by referring to Figs. 10, 11 and 12. Small inclusions also have an effect upon hardenability, as will be noticed from Fig. 8.

Metallographic examination of the center of each of these slabs did not show any difference in the amount of inclusions. Apparently the inclusions causing the change in hardening effect were submicroscopic in character.

In reply to the discussion by Dr. G. E. Doan may we say that we have been unable to find any correlation between the P-F characteristic and the gas content of the steel in question. Our oxygen analyses were made by the courtesy of Mr. E. C. Bain, and the nitrogen analyses were made by the courtesy of Halcomb Steel Company, and are as follows:

Sample No.	Type	Nitrogen Analysis Per Cent	Oxygen Analysis Per Cent
CA	1	0.006	0.007
AH	1	0.008	0.011
CK	1	0.008	0.012
6	3	0.007	0.013
1	3	0.10	0.014
CF	3	0.008	0.011

STEELS: THEIR SELECTION AND APPLICATION IN GENERAL SHOP PRACTICE

by FRANK J. ALLEN

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WHILE most companies which have grown beyond the personal management stage of their founders, have efficient and sometimes an elaborate purchasing department, this is often their only contact with the suppliers of their purchased needs. This condition applies to all purchased materials; here we are interested in it as it affects the buying of steel.

The selection of the proper steel as part of a product, as a tool for making the product or for plant maintenance, has a very definite bearing on the commercial success of the product. It must be recognized that before a purchasing department can even begin to function properly the material must be correctly specified, and on its receipt it should be checked to ensure that the specification is being observed.

It is asking too much to expect the purchasing department, manned by individuals of utterly different interests and training, to be abreast of the latest developments in the applications of the various steels and other materials. This, together with lack of organized guidance to the department, is seen in the swollen and unreasonable steel stocks held by many companies. While the brand names in the stock may appeal to the artistic sense of an enquiring executive, the fact that much of the steel, while intrinsically good, is unsuited to the purpose for which purchased, will hardly be soothing to his commercial acumen.

Quite a common condition to find is that the purchasing department may have to depend on several individuals for guidance on their steel purchases. With these individuals engaged in other organization and production functions the specification of steels, if their recommendations can be so dignified, is secondary. Their good intentions cannot replace study and investigation.

Sometimes purchasing departments drape themselves with much dignity and the fine policy of buying only the best. If buying the best meant buying the most suitable there would be little fault to be found. When, however, as so often happens, it means

buying the most expensive regardless of suitability, the policy appears as a cloak.

Many times the plan is to leave it up to the supplier. With reputable suppliers this plan may work, but not all suppliers even of the greatest integrity are equipped to function properly in this way. Even where they are, this plan seems to be an extension of the national pastime of passing the buck.

There should be close technical contact with the makers and sellers of steel, but a purchasing department engrossed in mercantile dealings in dollars and deliveries is not the right point for such contact. It becomes clearer every day that this matter of steel specification is one for the specialist. The function and responsibility for technical contact would belong to such an individual, as the Purchasing Agent has that of commercial contact.

The specification for a steel should be held as simple as possible in order not to unnecessarily restrict the field of supply. Where it is possible the standard specifications of the National Technical Societies and like bodies, such as the Society of Automotive Engineers and American Society for Testing Materials should be utilized. These specifications give a very wide range and are available from a wide field of supply. Sometimes, however, special requirements may render such adoption impossible. For general shop purposes this would be a rarity. It should always be remembered that a tool or machine part should be just good enough for the job. It is poor engineering to build a strength into a part which the nature of its operation makes certain shall never be called upon. The following three points should guide the study of specifications.

1. What are the ideal qualities of a steel to do the job.
2. Is such a steel available on the open market.
3. What would the effect be of using an available but less ideal steel.

Steel requirements may be divided roughly into two main classes:

1. Machinery steels used for machine parts.
2. Tool steels used to make a machine part or sometimes as the part itself.

The qualities needed may be possessed by a plain carbon steel, or may be produced therein by heat treatment, or by addition of alloying elements, or by both these methods.

For a reasoned selection of steel the significance of the presence or absence of the various elements in the steel must be understood. All steels contain carbon; it is of course the combina-

tion of the iron and carbon which constitutes steel. There are always present sulphur and phosphorus as impurities. The manganese and silicon which are always present function as deoxidizers. Of all the possible elements present in steel, carbon exerts the greatest influence. Its presence in greater or lesser degree will cause considerable variance in the physical properties of the steel and its reaction to heat treatment is also much greater than that of any other element present. The prime importance of the carbon content is basic to any study of steel. For each 0.1% of carbon added to naturally cooled steel up to 0.9% the physical properties are affected approximately as follows:

Tensile Strength is increased 9300 lbs. per sq. in.

Yield Point is increased 3900 lbs. per sq. in.

Elongation is reduced 4.50%.

Reduction of area is reduced 7.25%.

Above 1.00% the brittleness of steel increases rapidly and its use is confined to parts requiring high hardness and little toughness, as in tools. Further increase in the properties just noted are of course obtained by heat treatment.

Sulphur in steel causes "red shortness" or brittleness in rolling or forging. Its presence makes a free cutting steel.

Manganese adds to the tensile strength and acts as a cleansing agent. It counteracts the sulphur by formation of manganese sulphide in place of the lower melting point iron sulphide. This acts to cure the "red shortness" of the steel.

Silicon increases the strength of the steel.

Phosphorus causes brittleness or cold shortness. Small percentages increase the tensile strength and hardness of the steel. In the carbon steels there is in either the A.S.T.M. specifications or the S.A.E. specifications an available wide choice. The physical range covers a broad field. The tensile strength either in the as rolled, as cast, or the heat treated condition may be from 50,000 to 200,000 pounds per square inch. Similarly, the yield point may range from 25,000 to 180,000 pounds per square inch.

It is to the carbon steels that turn is made to satisfy the demand for a readily machinable steel for high production work. Bessemer Screw Stock is no doubt the most suitable steel where free cutting is the criterion. However, its indifferent physical qualities are well known. Where ductility and homogeneity are needed, it is to be avoided. The open hearth screw stock grade as covered by the S.A.E. specification 1120 while providing more reliable physical qualities loses considerable in the easy machining qualities. However, the high manganese, high sulphur, low carbon type steels as represented by the S.A.E. No. X1315 provide both reliability of physical properties and ease of machining.

This will be discussed further later. Here it may be said that if Bessemer screw stock be regarded for free cutting as 100%, the following table may be set up in order of machinability.

TABLE 1.
MACHINABILITY

1. Bessemer Screw Stock
2. High Mn., High S., Low C.
3. Open Hearth Screw Stock

The low carbon steels do not respond to heat treatment except carburizing. The regular screw stock grades are unsuitable for this purpose owing to the high sulphur content. A simple steel for carburizing is S.A.E. 1020, however, its drawback is its bad cutting qualities. Here again the high manganese, high sulphur, low carbon type enters in. This will carburize to the same case depth in 20% less time than the S.A.E. 1020 type. The case will be more uniform and the core tougher giving the finished piece longer life. Production may be increased in the heat treating of quantities by the time; or using lower temperatures with the same time, as for the S.A.E. 1020 grade, a saving would result in maintenance of carburizing pots and furnace.

A very great number and variety of machine parts are made from the medium carbon range steels, and given proper thought a still greater use could be made of them. By medium carbon is meant from .25 to .55 carbon. The parts for which these steels are used always have definite physical requirements. The S.A.E. steels No. 1030 to 1045 have proven their value in this class. Steels with a carbon content of from .20% to .30% are not affected in physical properties by heat treatment except to toughen. Their tensile strength will range from 65,000 to 90,000 pounds per square inch. With a carbon content from .35% up, there is a ready response to heat treatment. These steels are not readily machinable. While the importance of the machine part may render ready machinability of less importance than in the low carbon group, here again manganese steel will help out. The high manganese, high sulphur, medium carbon group will provide a type giving relatively higher physical tests, more uniform response to heat treatment, and much greater machinability. While there is a production demand for parts of higher quality, this type must be borne in mind.

The above mention of manganese steels among the straight carbon group is because of their frequent advantageous substitution for carbon steels at first cost so little above that of carbon steel. More specific mention may be made of them here. It used to be thought that 1% to 3% manganese in carbon steel of up to .50% carbon caused brittleness in the steel. This in spite of the fact that 1.5% to 2% manganese was used in producing non-shrinking,

oil quenching tool steels. Now there are produced a range of manganese machine steels, the low carbon type, with

C — .10 — .20%
Mn — 1.25 — 1.55%
P — .045% Max.

S. .08 — .13% and with increasing carbon content and varying manganese to C — .55 — .60%

Mn — .90 — 1.20%
P — .04% Max.
S — .055% Max.

The low carbon type meets the demand for a free cutting steel, capable of being carburized and free from inherent brittleness. Sulphur in steel has been regarded as a necessary evil. However, not only does the resultant manganese sulphide from the manganese present act to prevent red shortness in rolling, but it is a definite aid to machinability. Manganese tool steels are of the non-deforming type. The non-deforming property contributed by the manganese is a valuable feature in the low carbon manganese screw stock type steel. The importance of this steel for carburized parts of delicate section, to reduce warping in the quench to a minimum, is evident. Steels having a manganese content of from 2% to 7% are very special in their nature. Those with the manganese range from 8% to 13% are known as Hadfield's manganese steels. Their principal value is their resistance to wear under heavy pressure and severe abrasion. The steel work-hardens under cold working more than any other steel. Where the work is purely abrasive, and there is little or cold working, the steel loses its advantage. These steels, due to this work-hardening property, are commercially unmachinable; they must be ground. Nickel combines with iron in all proportions and the resulting alloys have a great variety of useful properties and applications. There is a pearlitic, martensitic and austenitic series of nickel steels, and of these the pearlitic are the most important commercially in ordinary shop practice. For tool steels the use of nickel is limited. Nickel is much used with other alloys in steel. Less than 3% nickel does not improve the physical properties of a steel in a worthwhile manner unless other alloys are present. More than 5% nickel gives a steel difficult to work and, unless the carbon is low, tends to the martensitic. In the pearlitic steels, that is, where the nickel does not exceed 5%, the nickel increases the hardness, tensile strength, and the elastic limit without impairing the ductility. In the as-rolled condition the effect of the presence of nickel can be noted without special heat treatment in the increase of the tensile strength and the elastic limit. Hence the use of nickel steels for large parts which cannot be heat treated easily. In heat treatment a milder quench may be used to produce a hardness obtainable only with a severe quench on a carbon steel of

equal carbon. This reduces quenching strains and distortion and there is an increase in depth of hardening in the quench. With their lower critical range, the nickel steels are annealed or hardened from a lower temperature than the carbon steels of equal carbon.

A readily available series of nickel steels is the S.A.E. 3½% nickel group. S.A.E. No. 2315 is primarily a carburizing steel. In carburizing the nickel retards the absorption of carbon. Both case and core are toughened by the restriction of grain growth during the carburizing and by the increase in the physical properties due to the presence of the nickel. Steel No. 2330 is most widely used where great strength and toughness are required. Steel No. 2350 has a field of service for large gears subjected to severe service.

Chromium stands second in extensive use as an alloying agent. It finds its widest application in conjunction with nickel or vanadium. Chromium has a great affinity for carbon and the formation of chromium carbide in steel proceeds until all available carbon is so used; only then does it go into solid solution in the ferrite of the steel. The hardness of chromium steel is partly due to this formation of chromium- and chromium-iron carbides. The hardness obtained does not affect the toughness given by the grain refining effect of the chromium. Chromium lowers the thermal conductivity of the steel. Thus it is important that heat be applied gradually in treatment to avoid setting up internal stresses due to unequal thermal expansion which might crack or deform the piece. Equally outstanding with the uniform hardness and depth of penetration of heat treatment caused by chromium's presence, is its effect on the magnetic properties and corrosion resistance of the steel. The following table may be quoted:

TABLE 2.

Chromium Content	Carbon	Property of Steel
½%		Deep uniform hardness.
2-3%	to 1.5%	Bearing steels, wear resistance.
3-4%	to .90%	Hot working die steels, permanent magnets.
12-16%	0.05-0.70%	Stainless steels.
10-20%	1.5-2.5%	Cold drawing dies. Very great wear resistance.

Vanadium may be termed the most powerful alloying element. It is used in smaller quantity than any other. Like chromium it is most effective when present with some other alloying element. Straight vanadium structural steels contain as little as 0.15% to 0.20% while in tool steels, high speed steel may contain up to 2½%. Vanadium intensifies the effects of heat treatment and its real use is in a heat treated part. For a given tensile strength the vanadium steels are more readily machinable than other alloy steels. Certain machining operations may be performed in the

heat treated condition. The most important industrial application of vanadium is in the *chrome-vanadium* steels. These are used in a great variety of ways for parts subject to severe service. They have excellent strength, toughness, resistance to wear and fatigue. Their machining and welding properties are good. For the various applications the necessary different characteristics are obtained by varying the carbon content. Generally the chromium is held at 0.80 to 1.10% and the vanadium at 0.15% min. with 0.18 desired. For parts requiring a very hard surface and subjected to shock the S.A.E. steel 6120 carburized has proven of value. The hard case and exceedingly tough core which are developed give the needed combination for shock and wear resistance. S.A.E. 6150 can be used over a wide field for machine parts which perform severe service. In the heat treated condition the tensile strength will run from 120,000 to 140,000 pounds per square inch and the yield point from 105,000 to 125,000 pounds per square inch. In this condition it is machinable and parts of shape difficult to heat-treat uniformly may be thus produced from the heat treated bar or block. Without developing extreme hardness, this steel has a very high resistance to wear as well as resistance to fatigue. This chrome-vanadium type has a naturally dense, homogeneous structure which develops a work-hardened case in service without loss of dimension. This case gives a wear resisting quality equal in many cases to a file-hard part. There is a very deep and uniform penetration of heat treatment. Chrome-vanadium steels in varying carbon content find many uses as tools.

The *nickel-chrome* steels resemble the chrome-vanadium series in many ways, yet they have important differences. They are used for structural parts requiring great strength and toughness, but have little application as tool steels owing to the lack of influence in this direction of the nickel. The nickel and chromium are used in structural steels in varying amounts but as a general rule the nickel content is about two to three times that of chromium. The carbon does not often exceed 0.50%. In nickel-chrome steels the nickel and chromium each contribute their individual characteristics and tend to minimize each other's disadvantages. The nickel toughens while the chromium increases hardness. The chromium tends to raise the critical while the nickel tends to lower it. The tendency to drag in machining of annealed nickel steels is somewhat offset by the presence of the chromium. S.A.E. 3120 is a case hardening steel which will develop the maximum core strength obtainable with the nickel-chrome steels and can be used where medium depth of case is satisfactory. For axles, shafts and other heavy duty the S.A.E. steel 3135 contributes strength, toughness and fair machinability. The increased carbon of the S.A.E. 3140 makes it susceptible to deeper penetration of the heat treatment and it is used for parts of heavy section.

Tungsten was one of the earliest elements used in alloying steel. It finds practically no use in structural steels since in the absence

of a high carbon content moderate quantities of tungsten have only a very slight effect on the mechanical properties. Tungsten is used in a great number of tool steels including high speed steel and in parts to resist high temperatures. It increases the hardness of a steel, is a grain refiner, and has marked effects on the magnetic properties.

Molybdenum resembles in some ways tungsten and in some ways vanadium in its effect on steel. It is used in moderate percentages in tool steels and magnet steels and still more in structural steels where in small percentages together with chromium the resulting steel is very much like the chrome-vanadium type. When present in small percentages molybdenum increases the elastic limit and shock resistance. Molybdenum steels in the higher range of hardness are easier to machine in a heat treated condition than other alloy steels of equal strength. The air hardening properties imparted to steel by molybdenum make it particularly valuable in the tool steel field. It must be clearly remembered in selecting an alloy steel that the physical properties of the steel in the heat treated condition must determine its use. Any advantage of the untreated alloy steel over the carbon steel are insufficient to warrant its use. The physical properties of untreated alloy steels are not reliable, depending as they do on the amount of work done on the steel at the mill and the final rolling temperature. It is sometimes necessary to substitute one alloy steel for another. The following list gives a possible substitution.

TABLE 3.

Specified S.A.E. No.	Possible Substitution S.A.E. No.
3120	2315, 2320, 3115, 6120, High Manganese Screw Stock
3135	2335, 2340, 3130, 3140, 6130, 6140
2315	2320, 3115, 3120, 6120, High Manganese Screw Stock
2330	2335, 2340, 3130, 3140, 6130, 6135
2350	2340, 3140, 6140
6120	2320, 3120, High Manganese Screw Stock
6140	2340, 2350, 3135, 3140, 6135
High Manganese Screw Stock	2320, 3120, 6120

Together with an understanding of the nature of the changes wrought in steel by the addition of the various alloying elements there must be a proper interpretation of the physical properties registered for them. In considering steel, just what is the importance of the tensile strength; what is meant by the number given to a steel and called its hardness number?

The *Elastic-limit* is a property of great importance in dealing with steel. It marks the practical limit of static-strength for machine and structural parts. However, one serious overload may change

the elastic limit of the steel and therefore the useful life of the part. For steel in a very brittle condition the elastic limit is a theoretical figure quite valueless.

The *Tensile Strength* is not as frequently of the direct importance that is sometimes supposed. In most cases the machine part will have failed before the full tensile strength is developed. The tensile strength does give a fair idea of the elastic strength of the steel and acts as an index of the resistance of the part to repeated stress or fatigue. It must always be remembered that in a long structural member or machine part, a compressive load may cause flexural action. Consequently the tensile stress set up in one side of the part will be the limiting factor of its strength.

The *hardness test* on a steel either by the Brinell, Rockwell or Scleroscope method is an easy test to make. The hardness number gives a reliable index of the tensile strength of the steel and thus of its fatigue or endurance limit. The wear resisting qualities of types of steel cannot be compared by their respective hardness readings, neither will the hardness number give an index of the steel's machinability. While limits of hardness exist to commercial machining, this is chiefly dependent upon the structure of the steel. There is a maximum repeated stress which a steel is capable of withstanding indefinitely. This measure of fatigue resistance is the *Fatigue or Endurance limit*. It has no relation to the ductility of the steel. Fatigue failure is sometimes incorrectly attributed to what is called "crystallization." Steel is inherently crystalline. Such failure is due to the gradual weakening and final fracture of the existing grain boundaries, not to the formation of any new crystalline structure.

Ductility will prevent shattering collapse when the elastic limit is reached and will, under occasional overload, tend to stop the formation of minute cracks which would become cause of progressive failure.

The *Elongation and Reduction in Area* give the best measure of ductility. Checking the physical properties of the steel to be used for a machine part, however, is of little use unless an equal check be kept upon the design of the part. Allowance must be made for the stresses set up at thread roots and bases of keyways. Wherever possible sharp shoulders, such as changes of diameter, should be replaced by fillets. The more generous the fillet the less will be the localized stress. Unless this is done active agencies of fatigue failure are set up. In checking design it must be remembered that the modulus of elasticity of steel is, for all practical purposes, the same for all types of steels, and is not affected by heat treatment. Within the elastic limit of the steel there will be uniform deflection of any given member regardless of the type steel used, or the heat treatment to which it is subjected.

When special circumstances require it, resort can be had to an alloy steel, and for the great majority of problems the answer can

be had from a relatively small number. A good carbon steel, of proper carbon content, and correctly treated can cover a vast field. The following tables 4 and 5 show a suggested set-up for standard machine steel specifications for general shop work.

TABLE 4.

<i>S.A.E. Steel or Equivalent</i>	<i>Tensile Strength—lbs. per sq. in.</i>	<i>Yield Point lbs. per sq. in.</i>	<i>Elong. in 2"</i>	<i>Red. in Area</i>	<i>Machinability</i>
Flange Plate ASTM Spec. S.A.E. 1112 S.A.E. 1120 X 1315 S.A.E. 1020 S.A.E. 1035 X 1350 S.A.E. 6120 S.A.E. 6140	55-65,000 65-75,000 50-60,000 70-80,000 90-110,000 125-135,000 145-160,000	25-30,000 35-45,000 30-35,000 40-50,000 55-60,000 115-125,000 120-135,000	.25% (in 8") 30-35% 25-35% 18-28% 20-30% 18-22% 15-20%	50-60% 45-60% 35-45% 35-45% 60-65% 40-50%	Poor excellent excellent excellent poor good good fair fair

TABLE 5.

<i>S.A.E. Steel or Equivalent</i>	<i>Uses</i>
Flange quality steel plate A.S.T.M. Spec. S.A.E. 1112 and 1120 X 1315 S.A.E. 1020 S.A.E. 1035 X 1350 S.A.E. 6120 S.A.E. 6140	Plate sprockets, rings and discs, various shapes not made in quantity but cut out with torch. Screw machine products may be surface hardened but not carburized. Lightly stressed parts. Screw machine products to be carburized. Small forgings, bolts, levers, brackets, carburized sprockets, pinions, and rollers. Parts where strength is not paramount. As for X1315 but lacks machinability. Good carburizing grade. Responds to heat treatment for higher physical properties. Machined parts and forgings requiring fair strength. Heavy duty shafts. Crank pins, piston rods, gears. As for S.A.E. 1035 but for more severe service. Better machining and response to heat treatment. Carburizing stock for severe wear and shock resistance. High shock and fatigue resistance. Uniform response to heat treatment. Highly stressed machine parts. Constant-mesh gears, worms, screws. Heavy duty gears and shafting. Cams, cam rollers.

TABLE 6.

The following table will serve as a guide to the relative machinability of some carbon and alloy steels.

<i>S.A.E. Steel</i>		<i>% Machinability</i>	<i>Type</i>
S.A.E. 1112 X1315 1120 X1350 6140 4130 2330 1040 1035 1020 6120 2345 3115 2315	 annealed annealed annealed annealed	100% 85% 80% 75% 65% 65% 65% 60% 60% 60% 60% 60% 60% 55% 55%	Bessemer Screw Stock High Manganese, High Sulphur Open Hearth Screw Stock High Manganese, High Sulphur Chromium-Vanadium Chromium-Molybdenum Nickel Machinery Steel Mild Steel Soft, Open-Hearth Steel Chromium-Vanadium Nickel Nickel-Chromium Nickel

As in the field of machinery steels, so is there available today a wide choice in tool steels. Here again all the particular circumstances of the job must be considered in selecting the steels and generalization is even more difficult and less helpful. For general shop practice tools may be divided roughly into three classes.

1. Cutting tools
2. Shearing tools
3. Forming tools

A cutting tool must resist wear, pressure, and heat. Mechanical means may be used to lessen the effects of all three of these conditions, namely, by alteration of the rake angle of the tool and the use of a coolant. The needed qualities in the tool are opposed to one another. Hardness to resist wear means brittleness, which will be susceptible to the pressure, while the heat generated will tend to destroy the hardness. It may be noted here that in all classes of work the tool set-up and the nature of the tool support, and the section of the part being machined as well as the nature of the material being machined, determine the necessary toughness of the tool to resist the shocks imposed upon it.

In the field of cutting tools carbon steel finds wide application. It develops a hard case and tough core but cannot withstand much heat. The addition of vanadium or chromium or both of these elements tends to toughen the steel and impart a deeper and more uniform hardness. Where service is not heavy and for small tools subject to abuse, carbon steel will answer the purpose. It is also much used for finishing cuts.

High speed steel is, however, the most generally used material for cutting tools now in use. The type most used is the 18-4-1 type, viz. 18% tungsten, 4% chromium, 1% vanadium. In addition to its cutting properties and ability to hold an edge at high temperatures it has a very high resistance to abrasion. A further development is cobalt high speed steel. The addition of cobalt increases to a remarkable degree the cutting efficiency of the tool. It increases the wear resistance and the red hardness. One of its chief applications is for hard, scaly material. However, this steel is subject to decarburization in any heating cycle and this may prove objectionable in cases where the tool cannot be ground after treatment. Tools made of any high speed steel should be well backed up.

Shearing tools may be either regular shears, punches, or blanking dies. For simple shears abrasive wear must be provided against and for this purpose, use can be made of the high carbon, high chromium types. Where it is not possible to properly reinforce the shear blade, toughness may be imparted to the blade itself by the use of the chrome vanadium type. For hot shearing, high speed steel, if backed up, will stand up well, otherwise resort must be had to the tungsten hot work steels.

Another field of use for carbon steels is for punches, the hard case giving wear resistance and the soft core the necessary toughness. In special cases where the punch can be held very rigid it is possible to use either a high carbon high chromium or a high speed steel punch.

In most cases, blanking dies call for a non-deforming steel. The manganese, oil hardening type answers this purpose. For dies of more intricate shape the high carbon, high chrome steels are used. Since both these types are oil hardening the warpage is small. Where the die is of intricate shape and with sections liable to cracking there can be used the cobalt, high carbon, high chrome steel which will harden very satisfactorily in air. Where the chief requirement is resistance to abrasive wear either a high speed steel, pack hardened, or a high carbon, high chrome steel will serve. For hot work either high speed steel or a tungsten hot work steel may be used.

In most cases shock is the principal factor to be considered in respect to forming dies. Toughness counts for more than heat resistance or wear resistance. This provides a field for the chrome-vanadium, chrome-nickel, and chrome-nickel-molybdenum steels. For deep drawing dies and cold forming, the characteristic hard case and tough core of the carbon and carbon-vanadium steels are most important.

A set-up for standard stock to cover general shop work is given in Tables 7 and 8.

TABLE 7.

DIE STEEL

Type	Use
Carbon Steel, C .80-.90	For toughness and hardness. Chisels, punches, shear blades, spring steel.
Carbon Steel, C 1.00-1.10	Where hardness is the chief requirement. For dies where first cost is important. Long life is not essential.
Manganese oil hardening, non-shrink type.	Dies not too complicated. Wear resistance needed.
High Carbon, High Chrome, oil quenching.	Dies of intricate shape.
Cobalt, High Carbon, High Chrome air quenching.	Dies of intricate section, danger of cracking in the quench.
Chrome-vanadium type.	For shock purposes.

TABLE 8.

CUTTING TOOLS

Type	Uses
Carbon type.	Moderate service. Finishing cuts.
High speed. Cobalt type	Very severe service. Great wear and heat resistance. Tool well backed up.
High speed. 18-4-1	Heavy service. Good life between grinds. Tool well backed up.
3-5% tungsten type	Finishing cuts of high degree.

Tool steels are used fairly widely for machine parts. Straight carbon drill rod, carbon 1.00 to 1.20%, and high speed drill rod are perhaps the most used. Ball and roller-bearing parts are subjected to wear without much shock and for them are used deep hardening tool steels. Chrome-vanadium steels of 0.50 to 1.00% carbon are used for grinding machine spindles, while the carbon-vanadium, carbon-molybdenum-tungsten and the high carbon-high chromium types are all finding duty as machine parts.

DISCUSSION

O. E. HARDER, *Chairman*

H. B. NORTHRUP, (*Director of Mineral Industries Extension, The Pennsylvania State College*): Mr. Allen's paper has been presented so well that it does not leave many opportunities for discussion of the main points presented. My attention has been directed to one or two points upon which I wish to comment.

These include the role sometimes assumed by a purchasing department in specifying analyses of steels, which role is frequently outside of their jurisdiction. To the two main classes of steel, namely, machinery steels and tool steels, I would add a third large class, namely production steels. This seems to be a very important classification and comprises not only a large tonnage but a diversified analysis. Silicon is not always present in appreciable amounts in production steels unless specified and controlled. I believe good practice should include a silicon specification of at least 0.10 per cent. In heat-treated and carburized steels the specifications for carbon and manganese should be so balanced that should a shipment be received within these specifications with both carbon and manganese on the high or on the low side, the heat treating effect of that combination will be within the limits expected.

With reference to carburizing steels, it has been my experience that S.A.E. 1020 steel is probably the world's worst selection for a carburizing specification. This is not alone because of its poor processing qualities—it has poor machinability—but more specifically because its carbon range is too wide and the upper carbon limit is too high. Mr. Allen states that addition of nickel alone under 3 per cent does not improve physical properties in a worthwhile manner. I know of many instances where the incorporation of 1.0 to 1.5 per cent of nickel to a straight carbon steel has made a difference between success of the part with the nickel added and failure of the straight carbon steel without the nickel addition. In ordinary carburizing runs up to about 8 hours at heat, the addition of nickel to a specification has no apparent retarding effect on carburization. Surprisingly good results are constantly being obtained in carburizing S.A.E. 1120 steel and also other steels with a specification including a similar high sulphur analysis.

WROUGHT ALUMINUM ALLOYS FOR STRUCTURAL APPLICATIONS

by E. H. DIX, JR.

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IN DISCUSSING aluminum alloys for structural applications, it is but natural that comparisons must be drawn between this, relatively speaking, new-comer in the field and the ancient and honorable iron-carbon alloy known as steel. I have no hesitation in drawing these comparisons because I have the honor of representing Pittsburgh, which is both one of the greatest steel centers in the world and the birthplace of the aluminum industry in this country. If by comparison with an old and thoroughly understood metal, engineers are able to more accurately appraise the field of usefulness of a new material, then humanity has been served.

The most economical metal for a given service will be the one that will ultimately survive in that service. Natural or artificial barriers will influence the economies of a particular application, but these effects are not lasting. Undoubtedly, if an industry is to grow its fields of application must be expanded and in this expansion other materials will be replaced with a resultant loss in tonnage. However, the introduction of a new material generally tends to increase the use of the commodities to which it is applied so that the net result is often an increased tonnage of the older metal. A further advantage of this sort of competition is the stimulation of research and development activities in all metals.

Those of you who may feel that aluminum is becoming a competitor of structural steel should reflect for a moment on the following figures. In 1929 the world production of steel was roughly 96,000,000 metric tons and that of aluminum approximately 280,000 metric tons; that is, there was roughly 350 times as much steel produced as aluminum, or expressing it in a different way, the total aluminum production was less than one third of one per cent of the steel production. Consequently, if all of the aluminum produced had been used to replace an equivalent amount of steel, the effect on the steel tonnage would still have been insignificant. Actually, only a relatively small proportion of this total aluminum was used in such applications.

Steel and aluminum have been close working team-mates from

the very first application of aluminum. In the early days of the aluminum industry, the high electrical conductivity of the metal suggested its application for electrical transmission lines, but this application seemed doomed to failure because the mechanical properties of the pure aluminum, required for maximum conductivity, were not sufficient to meet the service conditions. An invention of the late Mr. William Hoopes, combined the good mechanical properties of high strength steel wire and the high conductivity of aluminum by making a stranded aluminum cable with a galvanized steel core, thus starting A.C.S.R. transmission cables. (The letters stand for Aluminum Cable Steel Reinforced.) As a result, there are now considerably over 400,000 miles of steel cables, with their coatings of zinc, utilized in a field which would never have been available had it not been for aluminum. Numerous other instances might be mentioned but time prohibits.

In considering steel and aluminum for structural applications, there are three fundamental differences which now exist and which probably will be affected very little by future developments. These are (1) Difference in weight. Aluminum is approximately one-third as heavy as steel. (2) Difference in cost. By the very nature of the process required to produce metallic aluminum from the aluminum oxide obtained from the bauxite ore, the cost of producing aluminum is, and will probably always be many times as great as that of steel, unless some now-unforseen basic change in the method of producing aluminum is discovered. (3) Difference in Modulus of Elasticity. The modulus of elasticity of aluminum is approximately one-third that of steel and does not vary appreciably with the different alloys.

It is only through the full recognition of these three fundamental differences that economical and permanent applications can be found to increase the field of usefulness of aluminum as a structural material.

The specific gravity of aluminum alloys, of course, is determined by the specific gravity and amount of the alloying constituents. The aluminum-magnesium and the aluminum-silicon alloys are the lightest and the aluminum-zinc probably the heaviest, but except in special cases, as for instance, pistons of high-speed reciprocating engines, the specific gravity is generally not a controlling factor in the selection of an alloy. The modulus of elasticity, as we have said, is not appreciably affected by the various alloying elements. Other physical and mechanical properties may be varied over a wide range to suit the requirements of various applications.

Before discussing these applications, a few remarks concerning the effect of the difference in modulus of elasticity of steel and aluminum from a design viewpoint may not be amiss. The modulus of elasticity controls the "stiffness" of a given structure, assuming stresses below the elastic limit. If an aluminum alloy

is substituted, section for section, for structural steel, its weight will be 35.4 per cent of the steel, and it will be roughly one-third as stiff; that is, for a given load, it will deflect three times as much as steel. In many cases this increased deflection is of no importance, and may sometimes be an advantage. Assuming a square section for both metals, an aluminum beam having the same stiffness, as a steel beam, will weigh 61.4 per cent of the steel beam. For equivalent weights, the stiffness of the aluminum will be 265 per cent of that of steel. Obviously, the most economical use of aluminum is not obtained by maintaining the square section in increasing the area. If, instead, the depth of the beam only is increased, that is, the added metal is placed furthest from the neutral axis, then for equal stiffness, the aluminum will weigh only 51 per cent of the steel. You will notice that so far strength has not entered the picture. The above figures are true regardless of the strength of the two materials, remembering our assumption of stresses below the elastic limit.

Years of experience have demonstrated that the high strength aluminum alloys can be used to obtain a saving of about 50 per cent of the weight of structural steel.

The lower Modulus of Elasticity of aluminum alloys indicates a greater ability to absorb impact within ordinary working stresses.

The strength requirements of aluminum alloys for applications which have proved economically sound can best be discussed by reference to specific examples. For simplicity, tensile properties only will be used in comparing the various alloys. The shear strength may be taken as approximately 0.6 of the tensile strength. The endurance limit based on five hundred million cycles is from 5,000 to 20,000 lbs. per sq. in.

The alloy duralumin introduced in Germany about 1912 was extensively used during the War period for the construction of both heavier-than-air and lighter-than-air craft. The selection of this alloy composition is one of the most astounding metallurgical developments, because with the limited knowledge available at that time the inventor selected the three most useful alloying elements, in amounts and combinations so nearly ideal that today this alloy is still one of the most important structural aluminum alloy compositions. This alloy (4% Cu - 0.5% Mg - 0.5% Mn), designated 17S in this country, contains two elements, copper and magnesium, within the limits of solid solubility at the heat treating temperature and a relatively insoluble element, manganese. Iron and silicon are present as impurities. Magnesium combines with the silicon to form magnesium-silicide and copper with aluminum to form CuAl₃. These constituents make the alloy capable of effective heat treatment and manganese, as the compound MnAl₆ has a hardening and grain refining action. I mention these effects so that you may appreciate the complexity of the alloy

and recognize that whether or not the inventor understood the effect of these different elements as we do now, some twenty years later, he actually achieved a combination which for many purposes has not been greatly improved upon.

You may have noticed my continued emphasis in the above discussion on composition. While the composition has not changed appreciably, advances in methods of fabrication and particularly heat treatment have produced a structural material far superior to that used in early aircraft. The studies and theories evolved of the metallurgical characteristics of duralumin paved the way for what might be termed our modern "Precipitation Hardening Metallurgy." This had both a good and a bad influence on the acceptance of aluminum as a structural material.

Intergranular corrosion, now known to be common to most metal systems, was long emphasized in connection with the corrosion of duralumin because, owing to a lack of knowledge of the correct heat treating procedure, many early failures occurred through intergranular corrosion of the thin sheet used in aircraft. Today it is generally recognized that thin sections of duralumin require either a metallic coating of pure aluminum or paint protection to resist salt atmospheres. Sections $\frac{1}{4}$ inch or greater in thickness, properly fabricated, have sufficient corrosion resistance to withstand most conditions encountered in structural applications without protective measures. This is because in the very act of corroding, aluminum builds up a protective oxide coating which prevents further attack. In this process some of the aluminum is consumed, with resultant pitting which may cause difficulty in thin sections. For most structural applications, duralumin is given about the same sort of paint protection as structural steel.

The hot working characteristics of duralumin were early found to be unsuitable for forging, and in this country two alloys were developed; one containing copper and manganese as the principal alloying elements, and the other magnesium silicide. Both of these alloys forged very much more readily than 17S, and the former (25S) when fully heat treated develops about the same tensile properties. The latter (51S) develops only about two-thirds that tensile strength, with a higher yield strength and lower elongation. The heat treatment of these two alloys is performed in two steps. The first high temperature or solution heat treatment, followed by quenching, produces intermediate properties with considerable formability. The alloy 17S ages spontaneously at room temperature to full strength in about four days. Unlike 17S, these alloys require a final precipitation or artificial aging treatment to obtain maximum strength. A more recent development of 51S, designated A51S, contains a grain refiner and hence is more suitable for forged parts subjected to fatigue such as airplane engine crank cases. 25S-T is widely used for aircraft propeller blades.

The greatest economic advantage of the light weight of aluminum can be best realized in the transportation field. Its first application, as we have said, was to aircraft. Here weight saving is of supreme importance and cost of material a relatively small item. Alert and progressive designers have learned to take full advantage of the mechanical characteristics of aluminum alloys. In this field the continued urge has been for higher and higher strengths. This was first accomplished by cold working the material, generally by rolling, after final heat treatment. This raised the yield strength from approximately 35,000 lbs. per sq. in. to 45,000 lbs. per sq. in. with a smaller increase in tensile strength (58,000 lbs. per sq. in. to 61,000 lbs. per sq. in.) and a decrease in elongation (20% to 13%). Later a new alloy composition containing higher magnesium, about 1.5 per cent, was marketed in this country under the designation 24S (tensile strength 65,000 lbs. per sq. in., yield strength 43,000 lbs. per sq. in., and elongation 20%). Cold rolling applied after heat treatment to this alloy now represents the maximum in strength obtainable in a light weight alloy for aircraft (tensile strength 68,000 lbs. per sq. in., yield strength 53,000 lbs. per sq. in., and elongation 13%). This material covered with a thin layer of pure aluminum on both surfaces, marketed under the trade name, Al-clad 24S, in both the T and RT temper, was selected for the latest Douglas transports which represents probably the highest development in commercial aviation.

Aluminum has had a varied career in its application to automobile pleasure cars, buses, and trucks. You are probably familiar with early extensive use of aluminum crank cases, oil pans, gear boxes, and pistons. The pistons have represented a continuous, although fluctuating growth until today well over 50 per cent of the pleasure cars use aluminum pistons. The aluminum cylinder head, making possible higher compression ratios, better economies, and improved smoothness is in high favor today. Aluminum is a necessity in the power plants of modern buses and trucks.

In body construction, many of you remember the old aluminum bodies, both cast and wrought, for pleasure cars. The modern high capacity presses and improvements in forming qualities of steel have put aluminum out of this field. Instead, aluminum is finding increasing use in bus and truck bodies. The weight limitations on the highways have put a premium on light weight truck bodies, so that we find aluminum gasoline and milk tank trucks, aluminum bodies for large moving vans, coal and dump trucks, as well as many other applications.

For many of these parts, the high strength of the heat treated alloys is not required. The demand for alloys with properties intermediate between those of commercially pure aluminum and the strong alloys has resulted in the development of several new and commercially important alloys. The old alloy 3S, containing

about 1.25 per cent manganese, was found to be somewhat too weak for economical use. The next development was 4S, produced by the addition of about 1 per cent magnesium to the 3S composition. This alloy was stronger than 3S and found many useful applications, but is now being rapidly supplanted by the new alloy 52S, of slightly higher strength and better workability. It has very exceptional corrosion resistance, particularly to salt water conditions. It also has an exceptionally high endurance limit. These alloys, of course, are produced in various combinations of strength and ductility depending upon the degree of cold work.

It is not always economical or otherwise desirable to produce the required strength by cold working, and furthermore the stronger alloys of this type require high pressure for extrusion. These considerations led to the development of the alloy 53S, in which the hardening compound is magnesium silicide and the composition is controlled within unusually close limits to produce a combination of mechanical properties and high corrosion resistance previously unobtainable in a heat treated alloy.

As an evidence of the excellent corrosion resistance of these two new alloys, 52S and 53S, it may be stated that specimens $\frac{1}{16}$ inch thick showed no loss in tensile strength or elongation after one year's continuous exposure to the spray of a 20 per cent salt solution.

The general public is just becoming conscious of the desirability of lightweight steam and electric railway equipment. We have all become so accustomed to seeing massive steam locomotives and Pullman cars that there has been a tendency to accept weight as necessary to safety. Not long ago a well-known engineer stated that the only improvement that has been made in railroad car construction in the last ten years was the introduction of a slot for used safety razor blades in the Pullman cars. This, of course, is not so. The railroads are necessarily conservative and only adopt new developments after a thorough and conclusive trial. The new all-aluminum Pullman cars and the Union Pacific aluminum high-speed train, with which most of you are familiar, are the indirect result of some ten years' pioneering of aluminum in railroad construction. During this period approximately 15,000,000 pounds of aluminum has been used in nearly a thousand railway locomotives and cars and 1200 electric street cars, principally of the multiple-unit, rapid-transit type.

Much of the early development work was a mere piece-meal substitution of aluminum for steel in places where the light weight of aluminum would be most useful. As greater familiarity with the fabrication of aluminum was obtained, designs were changed to permit the most advantageous use of the material. The two all-aluminum Pullman cars exhibited at the Century of Progress represent the most advantageous use of aluminum alloys in the more or less conventional type of car design, whereas the

new light-weight high-speed articulated streamline train built by the Pullman Car and Mfg. Corporation for the Union Pacific System is a beautiful example of engineering design, particularly adapted to aluminum alloys.

In this latter design, full advantage was taken of the ability to extrude aluminum alloys into complicated sections. So accurately can these sections be extruded that the entire design is based upon interlocking sections producing a structure of minimum weight and deflection, with simple shop fabrication. The choice of the alloy for this application was based to no small extent on ease of extrusion and good forming characteristics. The alloy selected was A17S-T, containing 2.5 per cent copper, and 0.3 per cent magnesium. This alloy has a tensile strength of about 40,000 lbs. per sq. in., a yield strength of 22,000 lbs. per sq. in., and an elongation of 22 per cent. It obviously was not the highest strength alloy which could have been selected, but considering the other design factors and the method of fabrication, its strength was ample and its use economical.

It seems a long step from the dynamic applications that we have been considering to the economical use of aluminum alloys in a static structure, but this has occurred in the last year. The entire bridge floor system of the Smithfield Street bridge in Pittsburgh has been replaced with aluminum. The heavy steel and wood floor of this 50-year-old bridge had become inadequate for today's travel. By the use of 340 tons of aluminum alloy it was possible to reduce the dead load on the trusses by more than one ton per linear foot, thus permitting unrestricted traffic and adding an estimated 25 years to the life of the structure. The additional cost of the aluminum was small in comparison to the million and one-half dollars which would have been required for a new bridge. The alloy selected for the structural parts of the bridge floor had been undergoing development and test for a long period and its characteristics were well established. It was selected primarily for its high yield strength and good corrosion resistance. Typical physical properties of this new alloy, designated 27S-T, are: tensile strength 61,000 lbs. per sq. in., yield strength 49,000 lbs. per sq. in., and elongation 10 per cent. The shapes and plates required for this structure were of such size that their production would have been impossible had it not been for the foresight of the Aluminum Company of America in building the large modern structural mill at Massena, New York, and plate mills at Alcoa, Tennessee. At no other place in the world could this material have been produced.

Owing to the time limitation of our program, it has been necessary to limit my discussion to wrought aluminum alloys and to pass very briefly over the various structural applications. I have tried to give you a picture of the wrought alloy developments during the past lean years of the depression. A similar interest-

ing story could be told on the casting alloy developments and an almost equal period of time might profitably have been devoted to methods of fabrication including marked recent improvements in welding technique. I might also have dwelt at some length on the uses of aluminum for its chemical characteristics, such as tank cars for acetic acid, and other materials, corrosive to many metals, but without effect on aluminum. However, if I have succeeded in bringing to your attention some of the advantages of aluminum alloys for structural applications so that in the future you will think of these materials for applications where their use would result in a real economic gain, I will be well satisfied.

DISCUSSION

O. E. HARDER, *Chairman*

MR. LLOYD MORRIS, (*Pennsylvania Railroad Company Testing Laboratory, Altoona, Pa.*): The first application of aluminum alloys to the equipment of the Pennsylvania Railroad, in 1926, was in constructing the super-structure of 8 electrified cars. I believe this was the largest hot pressing operation on aluminum made outside of the Aluminum Company's shop. Some plates and sheets of 17ST were fabricated and heat treated at the plant of the Aluminum Company. The tensile strength of this material was about 61,000 pounds per square inch. Some parts were sent to the railroad shops and heat treated there after hot pressing between dies. A good many samples were tested after treatment at the railroad shop for strength and Brinell hardness in order to establish a relation between the two. When this relation was established, practically every part was tested on a Brinell machine. Since they thought it desirable to obtain long-time test results on specimens, some were laid away in the shops for a ten-year test. There has been seemingly no effect of time thus far with the exception of the possibility of a trend in the elongation and transverse strength.

About 70 locomotives have been built by the Pennsylvania Railroad with aluminum alloy super structures for the purpose of reducing dead weight. The only trouble encountered thus far has been corrosion of the roof sheets at the welded joints when improperly welded. New pieces properly welded withstood corrosion as well as the sheets themselves. The side and main rods for one locomotive were made of heat treated aluminum alloys with success, although the first rod put into service broke, due to improper design. In conclusion, I think that aluminum alloys are entirely satisfactory for reducing dead weight, the purpose for which they were used.

NORMAN I. STOTZ, (*Braeburn Alloy Steel Company, Braeburn, Pennsylvania.*): I wish to compliment Dr. Dix on his paper and

to point out that his statement with regard to the continuation of research unimpaired throughout the depression was significant. I think many other companies would be in better shape today if they had followed the example of the Aluminum Company in this regard, because the Aluminum Company today stands ready to produce shapes with new applications as the need arises.

J. W. FOSTER, (*Ingersoll-Rand Company, Phillipsburg, New Jersey—Presented by H. G. Day*): Mr. Dix has admirably summarized the present status of wrought aluminum alloys and has impressed upon us the rapidly growing application of these alloys to the realm of transportation.

Lest Mr. Dix's remarks leave too glowing a picture of the possibilities of aluminum we should remember the three fundamental differences (weight, cost and modulus) between steel and aluminum. The cost and low modulus are distinct disadvantages to which might be added: softness, poor corrosion resistance, high coefficient of expansion which presents quite a problem where elevated temperatures are involved. An additional disadvantage is the low yield strength, particularly when we consider that the yield strength of aluminum alloys is defined as the stress at which the stress-strain curve shows a departure of 0.2% from the initial modulus line produced.

The determination of the yield strength in this manner produces a considerably higher figure than is obtained by observing the "apparent" yield point on the stress-strain curve.

Despite these factors, the low specific gravity of the aluminum alloys makes their use commercially and economically mandatory in a rapidly increasing number of applications.

Mr. Dix states that the lower modulus of elasticity of aluminum alloys indicates a greater ability to absorb impact within ordinary working stresses. Resistance to impact is probably more dependent upon the ductility or toughness of a material than upon the modulus. Cast iron, for example, with a relatively low modulus, has practically no resistance to impact.

The user of wrought aluminum alloys these days has many combinations of alloys, heat treatments and tempers to choose from. In fact, with over forty such combinations in wrought alloys alone, the position of the user can be compared with that of a small boy in a candy store attempting to select the most tempting candies out of many: he has quite a job! (*Written Discussion.*)

HORACE C. KNERR, (*Consulting Metallurgical Engineer, President, Metallurgical Laboratories, Inc., Philadelphia, Pa.*): Mr. Dix compares wrought aluminum alloys with steel in structural applications. Such a comparison should include many factors, to be fair. The problem is quite complex. A more thorough analysis

will show that a greater share of the honors should go to steel, in its various forms, than Mr. Dix's paper seems to indicate.

STRENGTH-WEIGHT EFFICIENCY: Structural materials, where weight is considered, cannot be compared on the basis of either weight or strength alone. Pure aluminum, for example, is a light metal, but is a heavy material of construction because of its low strength. Aluminum alloys are sometimes referred to as the "lightest materials of construction." This is not true. A heat treated alloy steel of high strength is a good deal lighter, where it can be used efficiently, than any aluminum alloy available. This can be demonstrated in various ways, one means of comparison is the "strength-weight factor" obtained by dividing the tensile strength by the specific gravity of the material. If preferred, the yield strength, endurance limit, or other stress values may be used but most design is based primarily on tensile strength with due allowance for other mechanical properties which the designer deems important. A few typical strength factors follow:

STRENGTH-WEIGHT FACTORS

Tensile strength, thousands, divided by specific gravity.

	<i>Tensile Strength lbs. per sq. in.</i>	<i>S.W.F.</i>
Pure Aluminum	13,000	4.5
Iron, Ingot	40,000	5.5
Steel, cold rolled	60,000	7.7
Aluminum alloy 2 S-H	24,000	8.5
Aluminum alloy 17 S-T	58,000	20.5
Steel Alloy	160,000	20.5
Spruce Wood Aircraft	10,000	23
Aluminum Alloy 24 S-RT (Cold rolled strongest)	68,000	24.0
Steel (Alloy)	185,000	24.0
Steel Alloy, heat treated or corrosion resistance cold rolled	200,000	25.5
Steel (Very light)	250,000	32.0
Steel—Piano Wire	400,000	51.
Specific gravity—steel and iron	7.8	
“ “ pure aluminum	2.7	
“ “ aluminum alloy	2.8	

It will be seen that duralumin 17 S-T, at 58,000 pounds per square inch, compares with steel having a tensile strength of approximately 160,000 pounds per square inch. (Both are slightly heavier than aircraft spruce). The highest figure given for an aluminum alloy by Mr. Dix is 68,000 pounds per square inch for 24 S-RT which compares with steel at 185,000 pounds per square inch. In the aircraft industry alloyed steels for structural purposes are quite regularly heat treated to tensile strengths as high as 160,000 to 200,000 pounds per square inch and even as high as 240,000 and therefore much lighter. Piano wire is the lightest of all structural material.

The ratio of yield strength to ultimate strength for aluminum alloys is, in general, lower than for heat treated steels. On the basis of yield strength, steel is again the lightest material. The same is true of endurance limit or fatigue values. The endurance limits of aluminum alloys are rather low, the highest being 19,000 pounds per square inch for 52 S-H, which has a tensile strength of only 43,000 pounds per square inch. Alloy 24 S-RT has an endurance limit of only 14,500 lbs. with a tensile strength of 68,000 pounds per square inch which is a ratio of 21½%. The ratios for steels are considerably higher.

STIFFNESS: Where stiffness (elasticity) is the governing factor, a great deal depends upon whether the member is solid or hollow and whether its outside dimensions are limited. By increasing the depth of a beam, its stiffness can usually be greatly increased, or conversely, its weight reduced. For the same weight and shape, a solid aluminum section gains stiffness by its increase in depth. The material at the center or neutral axis has relatively little effect. If the beam (or column) is hollow, as it should be for greater lightness, then the wall thickness of the aluminum member will be three times that of the steel member. The stiffness and the weight will be about the same, the steel member being somewhat lighter for equal stiffness because its material is furthest from the neutral axis.

Where sheet is used, the greater thickness of aluminum alloy gives it greater stiffness over steel of equal weight. This greater thickness may also be of advantage in resisting denting or other detail failures.

FABRICATION: One of the greatest advantages of the aluminum alloys is their ease of fabrication. Aluminum alloys in the heat treated state can be quite readily cut with steel tools whereas alloy steels are difficult to fabricate in the heat treated condition and must usually be cut and formed in the annealed condition. (An exception is the case of the cold rolled, stainless steels, but these present special fabricating problems.)

WELDING: Where acetylene or arc welding is applicable in construction, steel has a great advantage over aluminum alloys which are practically limited to riveting. A well-designed, well-made welded joint is much more efficient, and therefore, lighter than a riveted joint, and often cheaper. In many instances, a steel structure may be heat treated after partial or complete assembly by welding. Heat treatment strengthens the weld as well as the parent metal and is common practice in aircraft construction. A welded structure gives greater strength-weight efficiency than a riveted structure, aside from the material.

The higher cost of aluminum alloys will often be offset by savings in fabrication cost, except in welded construction, where steel has the advantage. Aluminum alloys are obtainable virtually

from only one source, whereas alloy steels may be obtained from a great variety of sources and are, therefore, likely to be cheaper and more readily available in time of need.

CORROSION: When duralumin was first introduced into this country, extravagant claims were made regarding its corrosion-resistance. It was reported that samples had been exposed to sea water for six months or even two years without serious detriment. This resulted in some very disappointing experiences and a great loss both of money and of confidence in the material. Duralumin is seriously subject to corrosion, especially in thin sheet. The invention by Mr. Dix of the pure aluminum coating referred to probably saved duralumin as a material of construction in aircraft. This coating, having comparatively low tensile strength, detracts about 10% from the strength weight efficiency of the material.

The claims for corrosion resistance of the improved alloys are undoubtedly better founded than those for the earlier duralumin, nevertheless, corrosion-resistance should be taken under advisement in any given application until ample proof has been obtained in service. Certainly no aluminum alloy could be expected to compare with a corrosion-resistant steel in this respect. Both for ordinary alloy steels and for aluminum alloys corrosion is an important problem.

This brings us to the consideration of the corrosion-resistant steels. These are of two types, austenitic and martensitic. The former is made strong, and therefore, "light" only by severe cold work, such as rolling into very thin sheets or strips. In this way it can be given a tensile strength as high as 200,000 pounds per square inch with a very high yield point and excellent toughness or ductility. Its strength-weight factor exceeds that of the strongest aluminum alloys. It is very difficult to work or fabricate in this strong condition and special methods must be devised as in the Budd light train.

The other type which has received much less notice, but which may be destined for a very important field in structural applications, is the stainless iron type, martensitic, after quenching (eg. 0.10% C 16 Cr. I N). This can be fabricated in the annealed state like an ordinary structural alloy steel, and can then be given a tensile strength as high as 200,000 pounds per square inch with good elongation by quenching and tempering. The modulus of elasticity is about 30,000,000.

OVERLOAD OR CRASH: Despite theoretical considerations, it is doubtful whether duralumin will stand a crash as well as a good heat treated alloy steel. Experience in aircraft has rather indicated the contrary. Welded steel tubular construction is amazingly good. Fire hazard must also be considered. A steel structure has often passed through a serious fire without destruction where aluminum would have been a total loss. In one case an airplane

on the ground was completely destroyed by fire, except for the tubular steel structure. This remained intact and was again used and the plane flown in service. Aluminum alloy parts on the same plane were completely destroyed. Aluminum alloys are greatly weakened by even moderately elevated temperatures which would leave steel comparatively unaffected.

HEAT TREATMENT: Aluminum alloys are heat treated at relatively low temperatures, approximately 950 degrees F. as compared with relatively high temperatures for steel. This is an outstanding advantage in a certain sense, but on the other hand, the aluminum alloys are very easily ruined by slight over-heating and call for most careful and accurate temperature control. The fact that aluminum alloys are soft after quenching and attain their maximum hardness and strength only after natural or artificial aging, may also be very advantageous in certain fabricating problems. It should be noted that great progress has been made in the heat treatment of steel sections of very thin wall, such as aircraft wing beams which may be made of seamless steel tubing as thin as .035" and as long as 25 feet. These can be heat treated to 180,000 pounds per square inch or higher if desired, and held straight to better than 1/4 inch in their length.

CLOSURE: The writer having been one of the first to use aluminum alloys in this country, during his connection as metallurgist of the U. S. Naval Aircraft Factory, wishes to point out that he highly values and thoroughly appreciates the numerous important advantages of light aluminum alloys. These remarks are, therefore, made not in criticism of aluminum but rather in defense of steel whose excellent properties are in danger of being overshadowed by the effect of vigorous propaganda for aluminum alloys, or failure on the part of engineers and users to fully appreciate the undeveloped possibilities of light weight construction in steel. (*Written Discussion.*)

MR. DIX: There are several statements in Mr. Foster's discussion which are so much at variance with our experience that I would like to take a few moments to discuss them. He states that cost and low modulus are distinct disadvantages. Of course, this depends entirely upon the point of view. Granted, the lower the cost per unit weight or volume, the greater the advantage for a given material, but after all, cost is never merely the first cost of the metal. Advantages in ease of fabrication and yearly saving in operations determine whether or not a given material is economical. The added cost of aluminum is generally more than paid for in the first year of operation of trucks and busses. Aluminum alloys are proving economical in railroad applications. Aluminum in aircraft has been an economical necessity.

With regard to the modulus of elasticity, it is true, of course, that under some conditions of design, a higher modulus would be

desirable, but on the other hand, design based on the proper recognition of the modulus of elasticity generally results in a structure of about one-half the weight of that of mild steel and the increased bulk is often very desirable. Strange as it may seem, we know of applications where aluminum has been substituted for other materials in order to decrease the deflection of machine tools. A similar reduction in heavier metals could only be obtained by increasing the weight to a prohibitive figure.

To say that aluminum has poor corrosion resistance is distinctly inaccurate. The corrosion resistance of aluminum alloys varies, of course, with the composition of the alloys. As stated in my paper, we have recently tested sheet specimens $\frac{1}{16}$ inch thick which showed no loss in tensile properties after one year in the salt spray tank. Many aluminum alloys are being used because of their superior corrosion resistance. All of the commercial alloys have a corrosion resistance superior to structural steel. My statements are based on an extensive program of testing covering some ten years and including most of the other common metals for comparison. We have also much service experience to draw upon.

The question of yield strength is a matter which really should be discussed by testing engineers rather than metallurgists. The A.S.T.M. Standard Method for the determination of yield strength of aluminum alloys and other metals having similar stress-strain curves was used in obtaining the figures listed in my table. I think it is sufficient to say that designs based on the yield strength figures have proven entirely satisfactory.

We might argue for some time on the point concerning the effect of modulus of elasticity or ability to absorb impact. This much may be said, however, that in several wrecks which have occurred to aluminum interurban cars the damage has been judged by competent observers to be less than would have occurred in the case of steel structure.

In connection with Mr. Foster's reference to the number of wrought alloy combinations, our company, of course, does not recognize the necessity for as many as forty wrought aluminum alloys, but even so, the number of aluminum alloys is really small in comparison with the number of brass and bronze compositions and the various steels on the market.

Mr. Knerr's discussion rather gets over into the very controversial field of aluminum versus alloy steels. In my paper I attempted to use structural steel to form a correlation between recognized practice and the possibilities of the new aluminum alloys. A direct comparison between aluminum and the special steels for any given application is in itself a matter for a complete session and such a discussion would have prevented a presentation of pertinent information relative to the new aluminum alloys. Such a discussion is also more a question of engineering design than of metallurgy.

I note that Mr. Knerr felt that my discussion should include many factors in order to be fair. Likewise, I might say that his discussion advocating alloy steels omits the most serious disadvantage of these materials, namely, that the high strengths such as are cited in his table of Strength-Weight Factors can be utilized in most instances only by employing very thin sections. This, in itself, necessitates special and complicated designs which increase the already high cost of a costly material. It seems almost useless to enter into a discussion of other details because the engineering fraternity is apparently undecided regarding the relative merits of these materials and they are both being tried by very capable outfits. Time alone will give the answer. My guess is that, as usual, some sort of a compromise employing both materials, each where it is particularly adapted, will in the end give the most economical construction.

I am afraid that Mr. Knerr's picture of the corrosion resistance of duralumin is colored too much by aircraft experience where very thin sections are employed, and that he has been out of touch with developments in the past few years. Perhaps he would be interested in looking up, if not already familiar with it, the paper which I presented on the Corrosion Resistance of Structural Aluminum at the A.S.T.M. session last year. Further confirmatory information was presented at the Washington Symposium of the A.S.T.M. last March in a discussion which I presented for Mr. Finkelday's paper. I believe that this is to be printed. All of the evidence points to the fact that for sections of the thickness that would be used for structural purposes outside of aircraft, the corrosion resistance of even the less resistant aluminum alloys is sufficient if they are given protection such as is used for mild steel. In thin sections, Alclad 24S definitely shows less loss in properties than the cold rolled 18-8 material.

In my brief reply to his discussion, I took occasion to point out that he was the first man, as far as I knew, to publish the adverse effects of hot water quenching on the corrosion resistance of duralumin. This was certainly a major milestone in the development of aluminum alloys. I hope that, if opportunity presents, he will make us a visit at New Kensington so that we can show him some of the latest developments and our corrosion test results which lead me to speak with such confidence regarding the corrosion resistance of our structural materials.

alloys to resist these two process conditions, the problem is by no means solved and there is need for further and intensive research in this field.

DISCUSSION

DR. JOHN JOHNSTON, *Chairman*

FRANKLIN J. ALLISON, (*Crucible Steel Company of America, Pittsburgh, Pa.*): The author's paper is exceedingly interesting, not only in the results which he reports, but in the view which he gives us of the vast fields which are untouched in the matter of gases in metals at high temperatures.

In regard to hydrogen, he indicates that deleterious effects are due chiefly to two causes; first, to its embrittling effect which is well known to those who have had experience with pickling, and second, to its effect of combining with and removing carbon and other non-metals that are present in the metal.

The penetration or the removal of hydrogen from iron requires more time than is generally supposed. The importance of the time element was emphasized in an experience with 18% chromium tubing, which after pickling was quickly heated to 300°F. to remove hydrogen. Samples treated in this manner would crack on the upset test. It was thought that some other condition besides occluded hydrogen was causing the brittleness, until it was found that boiling in water for 3 hours eliminated the trouble. Thereafter, only "hard-boiled" samples were used for test. This experience was with samples tested fairly close to room temperature, but the literature shows that even at temperatures as high as 1800°F., as much time as three to five hours is required to remove all the hydrogen from iron powder,—and this is a condition in which a maximum surface of the material is presented.

In the case of most gases which enter metals, diffusion evidently takes place at a greater rate through the grain boundaries than through the crystals. Hydrogen in iron appears to be no exception, since carbon and other non-metallic elements in iron are expelled to the grain boundaries. They thus lie directly in the path of the greatest amount of incoming hydrogen and so may become readily attacked. The removal of these elements in gaseous form, such as CH_4 , H_2O , or H_2S breaks up the cohesion at the grain boundaries and causes disintegration of the metal. It would seem that this action could be minimized by producing material of small grain which would provide a baffle effect and a longer path for any gas traveling down the grain boundaries. It would also seem advisable to have the ferrous material in such condition of heat treatment that the carbides or other compounds would be well distributed throughout the grains and not concentrated at the grain boundaries. Thus a martensitic or sorbitic microstructure would be preferable to the coarse pearlite of a normalized steel. A further point in regard to decarburization by hydrogen, is

THE EFFECT OF GASES ON FERROUS MATERIALS AT HIGH TEMPERATURES AND HIGH PRESSURES

by HAROLD L. MAXWELL

Metallurgist, E. I. du Pont de Nemours and Company, Wilmington, Delaware

ABSTRACT

THE chemical processes developed during the last 15 to 20 years have placed additional responsibility on fabricators of chemical equipment to withstand the increased temperature and pressure encountered under service conditions. Perhaps the greatest change that has been made in the requirements for chemical equipment has been the result of processes dealing with high temperature, high pressure operations. This class of process might be divided approximately into two groups, one relating to hydrogenation and the other to high pressure synthesis, specially ammonia synthesis. The chemical processes involved in the first classification are concerned with the treatment of oils or fats at temperatures between 300° and 400°C. with hydrogen under pressures from 100 to 250 atmospheres. Under these conditions of temperature and pressure there is a tendency for the hydrogen to attack the metal resulting in embrittlement due to grain boundary attack. Metals used under these conditions ordinarily show a decrease in elongation and reduction of area with a moderate lowering in tensile strength. In the second case, where metals are used in equipment operating under ammonia synthesis process conditions the equipment parts might be exposed to temperatures in the vicinity of 500 to 550°C. and under gas pressures up to 1000 atmospheres or 15,000 lbs. The gas employed in this synthesis is a mixture of pure gases containing 75% hydrogen and 25% nitrogen. These are compressed and passed over a suitable catalyst, approximately under the conditions described, with the result that the molecule NH_3 , or anhydrous ammonia, is formed from the two constituent gases. The majority of metals suffer a rather rapid deterioration under the conditions of ammonia synthesis. Some of the early published results have shown that while a plain carbon steel, as for instance, .30 to .35 carbon, shows a rapid loss of properties under these conditions that by the addition of 2¼ to 3% chromium, with or without vanadium, the attack by the hydrogen is very greatly diminished. Although progress has been made during the past five years in the development of

whether the addition of elements such as chromium, tungsten, molybdenum, vanadium, titanium, and columbium which form stable carbides, would inhibit the hydrogen attack.

In reference to the author's results on the performance of the 18% Chrome - 8% Nickel steel in a hydrogen-nitrogen atmosphere at 550°C, it should be noted that this temperature is at the lower limit of the range where carbide precipitation takes place in the alloy, which fact may have had some contributing effect on the decrease in the mechanical properties. It would also be of interest to know the original condition of heat treatment of the alloy.

The poor performance of the 18% Chrome - 8% Nickel steel in the hydrogen-nitrogen atmosphere compared to what is generally accepted as the good performance of straight low chromium steel, raises some interesting theoretical questions. Is the poor performance due to the presence of nickel, to grain size, to grain boundary concentration of the carbide, or to the austenitic type of lattice?

One of the most noteworthy points in the paper, is the exposition of the complex reactions that occur in the chemical applications at high temperature. Even if there were readily available data on the effect of the individual gases, on the various alloy steels at elevated temperature, it would be difficult to apply such knowledge in such a case as the hydrogenation of coal or tar where, besides hydrogen, the reactions provide various gaseous hydro-carbons and sulphur gases. The author has not chosen to mention the effect of hydrogen sulphide, which is claimed in oil distilling and hydrogenation to be a particularly damaging gas at elevated temperatures.

From the paper it is evident that the selection of material cannot be laid along theoretical lines at present, and the best results will probably be awarded those efforts directed to drawing up the proper test procedure to allow empiric selection. (*Written Discussion.*)

T. HOLLAND NELSON, (*Villa Nova, Pa.*): I have read with a great deal of interest and pleasure Dr. Maxwell's paper. It is one that will be appreciated by those interested in the Chemical Engineering field, and is one more valuable contribution to the all too few available dealing with this very complicated subject.

The author is to be congratulated on bringing together in such a clear and concise manner the explanation of many results, which, worked out by individual investigators, have appeared to be conflicting. Dr. Maxwell has brought out very clearly that in dealing with gases under pressure at elevated temperatures it is by no means safe to assume that because a particular material works perfectly satisfactory at certain temperatures, concentrations and pressures that it can be taken for granted that this same material will prove equally as satisfactory if there be an alteration of any one of these three important factors. The writer, as a

pioneer in the field of corrosion resistant alloys, is confronted with this fact in all fields of corrosion resistance, and it is one I am glad to see brought out so clearly in this particular paper.

While I have had the opportunity of reading the text of this paper, the diagrams and microphotographs referred to were not attached and much of the information referred to in the paper is not clear without them. Therefore my comments are limited strictly to the text.

Earlier papers by J. S. Vanick, A.S.T.M., Vol. 2, 1924, and a more recent paper by John L. Cox, American Institute Chemical Engineering, June 15, 1933, which have a particular bearing upon the same subject and a review of the three papers at the same time gives one considerable room for thought. It is to be regretted, perhaps, that so far apparently no investigator has made a very exhaustive study of the behavior of the various metallic elements to bring out in simple form the effect upon them of hydrogen and nitrogen, or a mixture of these two gases at different temperatures and pressures. This is probably due to the fact that most investigators so far have been industrialists, usually attacking the problem from a practical point of view and having available only materials of probable utility for their particular purpose.

After several years of work in the field of corrosion, I am of the opinion that a great deal of valuable information is to be obtained from results on elements themselves, which, invariably, are indicative in measure of the properties they are able to confer upon iron as a base.

Looking at the available information to date, from this angle, we still find the subject somewhat confused. Vanick's work would suggest, in conjunction with that of John L. Cox, and others, that chromium figures very materially in retarding hydrogen infiltration, yet very little seems to have been done in the actual development of this thought along the lines of developing a definite curve over a range of chrome-iron alloys up to say as high as 30 percent chromium content. From a review of data available it would seem that with low temperatures and pressures comparatively simple and inexpensive chrome-nickel-vanadium-molybdenum and tungsten steels can be used with safety, and I wonder whether Dr. Maxwell's investigation enables us to foreshadow definite increase of alloy content in ratio to increase in temperature and pressures. Perhaps Dr. Maxwell's diagrams will throw a little more light on this subject.

Invariably we find because of the desirability of having low creep values with high physical properties at elevated temperatures, that the tendency has been to fall into the use of nickel-chrome alloys and one wonders whether it would not be advisable to investigate more thoroughly, with the object of finding out whether chrome or nickel really should be the predominating alloying element, or whether it proves to be essential that the two

together in iron, as a ternary combination, are essential.

It is extremely interesting to note from the work of Vanick that a material containing 16 percent chromium, 15 percent nickel, and 3 percent silicon apparently behaved remarkably well; in fact better than a material containing 80 percent nickel, 15 percent chromium, and 0.20 percent silicon. Here again, of course, temperature and pressure are different from those referred to by Dr. Maxwell, but are in this particular case comparable, being 500°C. and 1500 pounds pressure.

It is not unusual to find results accelerated with increased temperatures, concentrations, and pressures, but seldom have I found results reversed; by this I mean that if research investigations definitely show that chromium is an active agent in successfully retarding any type of corrosion, usually increased quantities can be adjusted to meet increased concentrations, etc. If, on the other hand, the retarding element be nickel, the same applies. Generally the use of ternary, or more complicated alloys are resorted to, either to obtain definite physical properties or for the purpose of facilitating fabrication. Generally I have found corrosion resistant alloys are capable of being classified into four groups, as follows:

- 1—Iron-Chrome
- 2—Iron-Nickel
- 3—Iron-Chrome-Nickel
- 4—Iron-Nickel-Chrome

One, of course, is quite willing to concede that other elements such as silicon, molybdenum, copper, titanium, tungsten, etc., have their particular advantages, but in the proper sense of the word the four groups above referred to are fundamentally the basis of practically all corrosion resistance. The behavior of tungsten, molybdenum, copper, and silicon are of particular interest in this field, and one wonders if Dr. Maxwell can inform us whether he has been able to draw any definite conclusions with regard to their use.

There is also one other rather important factor which, unfortunately, has to be very seriously considered: and that is, while the results of a definite research probably bring to light suitable alloy combinations for a specific purpose, it is not at all unusual to find that such alloy combinations are not commercially available. For instance, the research metallurgist today knows of several alloy combinations that would be highly desirable to the chemical engineer if they were available in the form he required. So that we are invariably called upon to modify the metallurgical ideal because of manufacturing and fabricating problems and we very often have to accept a "half-way" compromise.

From the writer's past experience he is of the opinion that dealing with straight hydrogen, the chrome series will probably be found of definite advantage in conjunction with Al., Cu. Si in smaller quantities, giving due regard to the fact that in a temperature range of approximately 450°F. to 550°F. these straight chromium alloys automatically become distinctly embrittled, though this brittleness is never apparent at the temperature but only on cooling. In this particular instance the ductility of such material can be readily restored by reheating to a temperature in the neighborhood of 1200°F., and in this direction one wonders whether Dr. Maxwell has done anything towards investigating the possibilities of a restoration treatment on alloys that have been subject to hydrogen absorption and infiltration. Assuming that carbon is removed as hydrocarbons, it is probable that restoration treatment would be of little, if any, value.

In passing over to the field of hydrogen and nitrogen mixed gases, experience in other fields might suggest that alloys of iron-nickel and chromium, in which nickel predominated the chromium content, would probably prove more advantageous than the straight chromium, and this thought would seem to be borne out by the information contained in the paper by John L. Cox referred to, wherein he refers to the excellent service of such materials as BTG, and possibly ATV, having approximately 60 percent nickel and 15 percent chromium on the one hand, and 35 percent nickel with 12 to 15 percent chromium on the other. In such cases I believe the reversal from a chrome predominant to a nickel predominant alloy is probably directly traceable to nitrogen and its well known nitriding effect in the particular range of temperatures dealt with in this paper.

I am particularly interested in the two instances referred to by Dr. Maxwell in his paper, being, respectively, that of an 18 and 8 low carbon, and the second instance where he refers to an alloy of 25 percent chromium and 20 percent nickel, and it is rather to be regretted that these two materials were not tested under identically the same conditions, so that there might have been a comparison throwing some light on the variations in chrome and nickel content. With regard to these two materials, it would have been extremely interesting to have available the varying carbon content as the effect of hydrogen progressed, and it would seem obvious from the tremendous change in physical properties shown by Dr. Maxwell that any of these corrosion resistant alloys susceptible to carbide precipitation must automatically be in a very weak state so far as infiltration by hydrogen is concerned, and one wonders what results might be obtained with titanium or columbium treated material of the same analysis.

While dealing with carbide precipitation in some of the lower alloyed austenitic steels, and with the knowledge that even in the presence of hydrogen chrome carbides are somewhat difficult to

break down, one would like to know whether Dr. Maxwell feels that carbides as precipitated in the 18 and 8 steels at the grain boundaries are likely to be substantially carbides of chromium, or double carbides of chromium and carbon. Apparently the facility with which they break down under the conditions outlined by Dr. Maxwell would have a tendency to shake one's faith in the theory of carbides of high chromium content, with possible chrome impoverishment in the vicinity, and I should be glad if Dr. Maxwell could throw any light on this very debatable point.

In conclusion one must again congratulate the author on so clearly pointing out to investigators the necessity of accepting very guardedly information relative to one set of conditions, and avoiding without further investigation the assumption that such results may be indicative of the performance of the same material when time, concentrations, and pressures vary. (*Written Discussion.*)

J. H. STOLL, (*Assistant Engineer of Tests, Bethlehem Steel Company, Bethlehem, Pa.*) (*Read by Mr. Foulkes*): One cannot help but be highly impressed by the paper which Dr. Maxwell has just presented. The thoroughness with which he has dealt with this problem which is of vast importance, not only to the chemical industry but also to the manufacturers of steel, is a tribute to his ability and I want to take this opportunity of expressing, not only for myself, but I feel sure for the steel industry at large, the appreciation and tribute due him for the exceedingly valuable information given us.

To be asked to give a discussion on this paper is a compliment I would be only too happy to accept, were it not for the fact that I am of the opinion it would be presumptuous for me to attempt to discuss the chemistry involved. Therefore, I offer you, first, my apologies and second, a few pertinent points relating to the manufacture of equipment, after the choice of the proper material, and present for your consideration certain phases involved in manufacture, all of which have a major bearing on the service and results which Dr. Maxwell has so efficiently studied.

It has been my privilege to be associated with the manufacture of considerable equipment, of the type used in the modern methods of chemical synthesis, to hold these gases at pressures of 15,000 lbs. per square inch at temperatures ranging up to 600°C. These problems of operating conditions are only magnified when it is appreciated that equipment of this type may run up in size to a forging weighing 245,000 lbs. To produce a forging of this weight, it requires an ingot weighing considerably more than 400,000 lbs.

It is appreciated that Dr. Maxwell has not in any respect exaggerated the necessity of the greatest uniformity possible to obtain in order to minimize the attack of gas on the vessel. Uniformity in the vessel, subject to this attack and capable of with-

standing the necessary pressure and temperature of operation, does not refer only to the finished machined wall section, falling within the allowable tolerance, but covers analysis, grain size, structure, freedom from strains, and numerous other factors going back to the first operations of manufacture. The effect of the lack of complete deoxidation was noted in the test made on copper, and although its effect in this test was far more outstanding than would be noted in steel manufacture, it is, nevertheless, an important factor. The effect of deoxidation is given due consideration in the melting of the steel and the open hearth furnace is carefully controlled and the proper slag balance maintained to produce a thoroughly deoxidized clean steel.

The attack of the gas on metal of different compositions is given consideration when the ingot is poured and first mechanically worked. Segregation in an ingot is inevitable, as it is fundamental in the laws of solidification, but it can be minimized, and, being a known quantity, can be reckoned with. Not only is the highly segregated metal in the top of the ingot discarded, but the center segregated metal is removed by trepanning or boring, with the result that the wall of the finished vessel does not contain the segregated center of an ingot, but is composed of only the most desirable portion of the ingot.

Dr. Maxwell has mentioned fissuring around the grain boundaries, which again brings up the thought of uniformity. Although we are accustomed to think of heat treatment as the application of temperature to a piece for the breaking up of a coarse grained structure, to a certain microscopic fine grained structure, together with the development of the desired physical properties, we can rightfully include the application of mechanical work at temperature to this recrystallization, which we know and speak of as forging. Uniform hot working, where each section of the wall of the vessel receives the same amount of work, serves not only to bring the metal into the desired form or shape but aids materially in establishing a finer grain by mechanical means. The ingotism or coarse grain of the cast structure is broken up, and we can therefore consider this operation somewhat as a conditioning treatment for subsequent final heat treatment operations.

The strains resulting from the forging operation must be relieved, and the forging structure further broken up and refined, to obtain the necessary uniform fine grained structure required. The heat treatment to be given the forging is selected with consideration to the chemical composition, the size and shape of the forging, the physical characteristics desired, and the type of service to which it will be subjected. Methods of heat treatment, which are considered standard for small uniform sections of material, are generally entirely unsuitable for larger forgings, it being necessary in many cases not only to alter the treatment tem-

peratures but also to change the entire scheme of heat treatment operations. It is also a well-recognized fact that combination of physical characteristics which are obtained on small sections of material, of a given composition, cannot be approached in heavier cross sections. Generally speaking, the cross sectional area and uniformity in dimensions, determine the type of treatment which is safe and effective.

With the ever-increasing demands for more exacting performance, forgings have been chosen to answer the need when the most exacting service is imposed on material. The uniformity and homogeneity which it is possible to obtain, through the operations which have been mentioned, are ample explanation of why the steel manufacturing industry has been able to keep in step with the rigid demands of the chemical industry. The steel industry at large is frequently criticized for its inability to immediately offer a solution to many manufacturing problems. I intimated earlier in my remarks that Dr. Maxwell has told us things about our materials which we did not know. We acknowledge this and further add, that we welcome all information developed in service pertaining to materials used in the manufacture of equipment. The problems of the steel manufacturing industry are in themselves of such a magnitude, that it is virtually impossible for the steel manufacturer to intimately know, and develop to the fullest extent, the personalities of materials used in the many different types of service.

Only too frequently is the manufacturer placed in the position of the middle man, in that they have virtually no hand in the engineering development of the equipment to be used, and little or no knowledge of the details of the service operations. Users of this large type of equipment are fast recognizing the advantage, to both themselves and the manufacturer, of discussing freely and frankly the designing and engineering features of their equipment. This reflects both in a saving to the consumer and in the production of the highest possible quality product.

In conclusion, we therefore wish to emphasize the closing remarks of Dr. Maxwell, to the effect that the answer in obtaining the most satisfactory material for these exacting service requirements, is very often best found by selecting the most likely looking possibilities among the many materials available, and subjecting them to test conditions, to develop their full value for the service imposed. Dr. Maxwell's contribution to the industry, on steel behavior in the drastic service he has discussed, is most commendable and we trust it is only the forerunner of much additional information, which will be mutually beneficial to both consumers and producers. (*Written Discussion.*)

PUBLICATIONS OF THE MINERAL INDUSTRIES EXPERIMENT STATION

Research results of the Experiment Station are disseminated through the following publications: (1) Bulletins, which present the developed results of experimental studies of a comprehensive problem, and the proceedings of technical conferences, (2) Information Circulars, which present in more or less non-technical language, the results of studies which are given in greater detail in other publications, statistical data, or pertinent information gathered from other sources. (3) Technical Papers, consisting of bound copies of papers published in scientific journals, of progress reports, and of results of experimental studies representing isolated phases of research, later to be published in bulletin form.

A few of the publications are listed below. These may be obtained from the Director of Mineral Industries Research, The Pennsylvania State College, State College, Pennsylvania, at the price stated.

Bulletins

15. Proceedings of the Third Pennsylvania Mineral Industries Conference, *Coal Section*, held at The Pennsylvania State College, November 10-11, 1933. 57 pages with 6 illustrations. Price, 50 cents.
16. Proceedings of the Spring Meeting, American Petroleum Institute, Division of Production, Eastern District, held at The Pennsylvania State College, April 6-7, 1934. 69 pages with 11 illustrations. Price, 50 cents.
18. Proceedings of the Fourth Pennsylvania Mineral Industries Conference, *Metallurgy Section*, Pennsylvania Inter-Chapter Meeting of The American Society for Metals, held at The Pennsylvania State College, May 4-5, 1934. 138 pages with 36 illustrations. Price, 50 cents.

Circulars

3. Opportunities in Pennsylvania's Basic Industries, 1932. 36 pages with 38 illustrations. Free.
4. Basic Trends in Mineral Industries Education, by Edward Steidle. 1933. 13 pages. Free.

Technical Papers

1. Variations in Microstructure Inherent in Process of Manufacturing Extruded and Forged Brass, by Ogden B. Malin. 1932. 9 pages with 6 illustrations. Free.
10. Observing Formation of Martensite in Certain Alloy Steels at Low Temperatures, by O. A. Knight and Helmut Müller-Stock. 1934. 7 pages with 4 illustrations. Price, 10 cents.