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# STEEL *and Its* HEAT TREATMENT

*Volume I · Principles*

STEEL AND ITS HEAT TREATMENT

BY

D. K. BULLENS

AND

THE METALLURGICAL STAFF OF THE  
BATTELLE MEMORIAL INSTITUTE

*Fifth Edition*

**VOL. I. PRINCIPLES.** The principles which relate to the heat treatment of steel in general, whether it be plain carbon or alloy steel. 489 pages, 5 $\frac{7}{8}$  by 9. 262 figures. Cloth.

**VOL. II. TOOLS, PROCESSES, CONTROL.** The practical application of heat treatment. 293 pages, 5 $\frac{7}{8}$  by 9. 118 figures. Cloth.

**VOL. III. ENGINEERING AND SPECIAL-PURPOSE STEELS.** A discussion of engineering carbon, simple alloy, complex alloy, and special-purpose steels. In preparation.

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*and Its*  
HEAT TREATMENT

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THE METALLURGICAL STAFF  
*Of the Battelle Memorial Institute*

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IN MEMORY OF

*Albert Nelson Bullens*

AND

TO OUR FRIEND

*J. A. Doyle*





# PREFACE

## To the Fifth Edition

Each decade brings advances in the understanding of the phenomena involved in making and treating steel. As understanding increases, ability to control the phenomena increases. The past decade has been no exception; indeed, due to huge production of heat-treated steels for implements of warfare, it has covered a period rich in experience.

Bits and pieces of scientific and technical knowledge steadily appear in appropriate publications. It is the task of such a book as this to pick out, from this accumulated store, the new, and the especially illuminating findings and string them together in an arrangement aimed to aid the practical heat treater.

So much that is new or illuminating has come up in the past few years that another revision of this book is called for. Major revisions of the 1915 edition were made in 1927 and 1938, nor is the present to be considered a "final" revision. There is no such thing as finality about such a topic; further and greater progress is inevitable. From time to time, however, it is worth while to record some decrease in our ignorance.

As time passes, the line of demarcation between the scientist and the technician grows fainter. Cookbook recipes for heat treatment do not cover the contingencies that arise in practice, whereas, with an understanding of the findings and the reasoning of the scientist, the technician can ably deal with most of these contingencies. The better grasp he has of the "why" back of the phenomena that occur, the better he can control them. Much of the "why" of behavior in heat treating is mapped out in "S curves" and "Jominy curves" which have been available for only a few years but, in spite of definite limitations as to applicability, are becoming a new but universal language of the heat treater and of the student of metallurgy.

There is no dearth of current reading matter about heat treatment; the difficulties are in keeping up with it all and in grasping the implications of some of the highbrow writings. Nor is there a dearth

of good textbooks, though some are written for the college student who has yet to meet practical problems and therefore naturally phrase the discussion from the academic point of view. The same facts and principles are set forth in different words and with varying emphasis by different authors, a helpful condition, since reiteration from more than one point of view aids in fixing the facts in the mind of the reader. Differences in interpretation of the same facts are thus brought out, and dogmatic statements by any one author may be taken only for what they are worth.

Not all heat treaters are privileged to spend four years on a college course in metallurgy, nor is it necessary that they should. Reading, plus careful observation of steel's behavior in practical heat treatment, should suffice. Excerpts from a recent pamphlet\* are worth quoting in this connection.

#### *No Formal Education*

At the bottom of the middle column of page 605 of the *Columbia Encyclopedia* you will find this brief biographical sketch:

*"Faraday, Michael 1791-1867, English scientist. Receiving no formal education, he acquired much knowledge which was to serve him later during his apprenticeship to a bookbinder. In 1813, he was appointed assistant to Sir Humphrey Davy at the Royal Institute in London. He is known especially for his discovery of electromagnetic induction, his study of the transformation of energy from mechanical to electrical, which led to the later discovery of the electric generator, and for his formulation of the laws of electrolysis. He wrote Experimental Researches in Electricity (3 Vols. 1839-55) and Researches in Chemistry and Physics (1859)."*

Dry as it is in expression, this condensed life story of a great scientist stirs us strangely. We reread the opening statement: "Receiving no formal education, he acquired much knowledge . . ."

This brings up a question we have been pondering lately: Is this generation *overeducated?*

Do we spend so many years cramming into our minds the ideas and theories of other men that we have neither time nor inclination to do any observing or thinking on our own account? Are we so awed by the authority of textbooks that we dare not question their theses and theorems? Is education by the direct observation of nature and people and the processes of life entirely outmoded? *Worse, are we perpetuating erroneous ideas and theories?*

Dr. Gerald SeBoyar of New York University says, "Education is merely training the mind to do individual thinking." Admittedly that is what it ought to be.

\* Brake Shoe Private Wire, No. 14, 1946. American Brake Shoe Company.

But along comes a Faraday: "Receiving no formal education, he acquired much knowledge . . ."

It might be a good idea for all of us, whether we are college graduates, or had no schooling beyond the sixth grade, to take a fresh look around us, with curious and questioning eyes.

If we begin with certainties, we shall end in doubts; but if we begin with doubts, and are patient in them, we shall end in certainties.—Francis Bacon.

It is in this questioning spirit about the ideas and theories of heat treatment that this revision has been made. Theories are necessary for clarification of principles, but more important are the experimentally established facts.

The previous revision dealt with principles and processes in Vol. I, with primary consideration of carbon steel and postponement of major discussion of alloy steels to Vol. II. In the light of present thinking, such a division is an artificial one, for the principles are often more clearly illustrated by alloy than by carbon-steel examples.

Important new bits of information about, and new methods of evaluation of, what goes on in heat treatment had to be included in the discussion of principles. The inclusion of the new matter would have made too bulky a volume, so the present revision will appear in three volumes.

This volume deals with the definitions, terminology, and fundamental concepts necessary for discussion of heat treatment; with the limitations and uses of tests applied to evaluate suitability of steel for heat treatment, and suitability of the heat-treated product for engineering use; and with the principles underlying the various types of heat treatment, whether of carbon or alloy steel.

Volume II will deal with applications of the principles and with the tools of heat treatment, such as furnaces, devices for local heating, salt baths, controlled atmospheres, cyaniding, carburizing and nitriding; quenching media; prevention of cracking and distortion; the processes used to put steel in proper condition for machining; the relation of heat treatment to welding; and the heat treatment of steel castings, malleable iron and cast iron. That is, it is aimed to cover the practical application of the principles described in Vol. I.

Volume III will discuss constructional and alloy steels, including carbon, mild alloy, N.E., SAE and tool steels, from the point of view of their selection (and of the utilization of alternate steels), on the basis of amenability to the principles and processes of heat treatment and of consequent suitability for various engineering uses.

In all, the aim is to present information of practical utility, and to show the limitations as well as the possibilities of present methods, equipment, and tests. A realistic attitude of mind toward many of these, coupled with an understanding of basic principles, goes far toward avoidance of blind adherence to cookbook recipes, outmoded methods, and those that ought to be outmoded.

## Acknowledgments, Fifth Edition

Those members of the Battelle staff who helped in the previous revision, plus S. L. Hoyt, H. M. Banta, A. R. Elsea, and others, have been equally helpful with this. As before, H. W. Gillett is responsible for the final selection and arrangement.

The publications of the American Society for Metals have continued to supply much of the basic material. Permission of the ASM, the AIME, and various technical journals for use of material published by them, is gratefully acknowledged. The sources of information are shown in the bibliography.

Especially thanks are due those authors who presented significant information in plots whose reproduction brings out the point under discussion, so that the compiler needs only to place them in logical order to knit together these separate strands.

D. K. BULLENS  
CLYDE E. WILLIAMS, Director  
Battelle Memorial Institute

# PREFACE

## To the Fourth Edition

With the exception of a chapter on nitriding, added in the 1935 printing, this work has not been materially revised since 1927. Much new information has become available, and many of the old principles have received fresh verification in the last decade. Hence, for the volume to continue to serve its purpose, revision in the light of present knowledge became imperative. Since pressure of other work prevented the original author from taking an active part in the revision, he invited Battelle Memorial Institute to carry it out.

Battelle Memorial Institute considered that the previous editions of the book had served as an unusually effective link between theory and practice, largely because of the sound and practical point of view adopted in emphasizing the importance of the human element and in dealing with the various heating processes as a broad engineering problem. Believing that this point of view deserves perpetuation, and that to aid in putting continued emphasis upon it is in line with the primary purpose for which the Institute was founded, that is, helping to put science to practical everyday use in industry, Battelle accepted the invitation.

Among features in which major advances have been made since the last edition are, first and most important, an understanding of the significance of the grain-coarsening tendencies of steels and developments in their control; the use of controlled atmospheres; the rise of high-yield-strength low-alloy steels, insulating refractories for furnaces, forced circulation for more uniform heating in tempering, gas carburizing, and other processes; the use of gas-fired radiant tubes for large furnaces at moderate temperatures; the use of copper as an alloying element in steel; the normalizing instead of annealing of steel castings; and, in general, a clearer conception of the mechanism of heat treatment and of the specific effect of all the alloying elements.

These, together with the increase in special "tailor-made" steels, of too great importance to pass by entirely, necessitate the introduction of new chapters and, in order to conserve space, the condensation or elimination of other matter.

Even with the elimination of such matter as could be spared, the material that had to be added brought the bulk to too large an amount for a single volume. In the division into two volumes, those principles, processes, and fundamentals of control which relate to heat treatment of steel in general, whether it be plain carbon or alloy steel, were placed in Vol. I, and the discussion of engineering carbon, simple alloy, complex alloy, and special-purpose steels, in Vol. II.

Practically everything in these volumes has, naturally, been recorded in other books and in periodical literature. The excellent monographs in the *Alloys of Iron* series, the *Book of Stainless Steel*, and other works of like completeness, including some put out by those who have a specific alloying metal to sell (which are valuable even if they do "put the best foot forward"), deal with each alloying element in a degree of detail possible in ample space. Valuable handbooks treat separate topics in tabloid form. The reader is especially urged to supplement this book by study of the appropriate sections of the *American Society for Metals Handbook*. Symposia on special topics, held by technical societies, supply compact information in especially convenient form. Back of these compilations are thousands upon thousands of articles in which the information was originally presented. This very profusion, although of immense value when all available information on a specific point is necessary, is a drawback when what is desired is a broad picture of the state of the art of the heat treatment of steels and an understanding of the fundamental principles involved, a picture which will enable the reader to apply the summation of known facts to the problems he personally faces.

Hence, the effort in this revision, as in the previous editions, is to present the whole topic from a consistent unified point of view, in which the need of the practical man for pertinent information is never lost sight of.

### Acknowledgments, Fourth Edition

The acknowledgment to those who aided in the preparation of the previous edition, J. D. Cutter, J. A. Doyle, C. N. Dawe, J. V. Emmons, H. S. Foote, H. J. French, L. Garratt, A. W. F. Green, H. L. Green, M. A. Grossman, J. E. Harper, C. M. Johnson, A. H. Kingsbury, E. Larned, F. B. Lounsberry, A. F. MacFarland, J. A. Mathews, A. H. Miller, H. F. Moore, O. K. Parmiter, N. S. Stansel and T. W. Wickenden, made in the preface to the previous edition, is as heartily made

here, since their co-operation helped build the framework of the present volume.

Added unpublished information or helpful comment on specific topics has been received from J. A. Doyle, A. J. Herzig, J. V. Emmons, C. H. Wills, and G. V. Luerssen, besides those to whom specific acknowledgment is made in the text.

Naturally, a primary source of information on advances in the heat treatment of steel is the publications of the American Society for Metals. The co-operation of the ASM in permitting the use herein of any material appearing in any of its publications has materially facilitated the revision.

The Society of Automotive Engineers has similarly permitted the use of its data on SAE steels, and other data have been taken with permission, from the publications of the American Institute of Mining and Metallurgical Engineers, of the American Society of Testing Materials, and from *Metals and Alloys*.

Permission for use of material appearing in *Nickel Alloy Steels*, *Vanadium Steels and Irons*, and *Bethlehem Alloy Steels*, published, respectively, by the International Nickel Company, Vanadium Corporation of America, and the Bethlehem Steel Company, has likewise been an important aid.

The first draft of the revision, drawn up by H. W. Gillett, Battelle Memorial Institute, was read by the original author and by S. Epstein, Bethlehem Steel Company. The helpful comments of both have been utilized. The detailed study and constructive criticism of the whole draft by O. E. Harder and C. H. Lorig of Battelle deserve especial acknowledgment. Among the other members of the Battelle staff who have supplied information and criticism on particular chapters or topics are R. W. Dayton, H. C. Cross, B. W. Gonser, H. W. Russell, R. A. Sherman, and C. E. Sims. The index was prepared, references checked, and many tedious details attended to by Miss T. R. Reinberg. H. W. Gillett carried out the final selection and arrangement of the material.

All temperatures mentioned herein are in the Fahrenheit scale, and the F is omitted, that is, 1000° means 1000° F. For those who think in centigrade, a conversion table is given on p. xx. Mechanical properties are given in U. S. units, and, to save space, are abbreviated; for example, instead of "Tensile strength, 100,000 lb/in.<sup>2</sup>, Yield strength, 80,000 lb/in.<sup>2</sup>, Elongation, 25 per cent in 2 inches, Reduction of area, 60 per cent, Izod impact 55 ft-lb," the notation "Tensile 100,000, Yield 80,000, Elong, 25%, R.A. 60%, Izod 55" will be em-



ployed. Similarly the name of an alloying element which recurs frequently in a discussion will not be spelled out; the chemical symbol will be used.

“Yield strength” is used throughout in place of “elastic limit” in accordance with the terminology of the American Society for Testing Materials.

The sections on nitriding are based upon the chapter written by V. O. Homerberg for the 3d printing of the 3d edition.

D. K. BULLENS  
CLYDE E. WILLIAMS, Director  
Battelle Memorial Institute

# PREFACE

## To the First Edition

Modern heat treatment should be considered as an art or trade, since it certainly requires knowledge, skill, and judgment for its proper performance. These, in turn, necessitate at least some knowledge of heat, of steel, and of the effect of heat on steel. And all three factors are linked together by the "human element." The author has therefore endeavored to bring together the theoretical and practical sides of the general subject of steel and its heat treatment in such a manner as will, he hopes, be understandable by that "human element."

The theoretical or scientific side of heat treatment is not affected by a consideration of many elements which enter into the art or practical application of metallurgical principles, but unless performance in the shop compares favorably with the standard set in the laboratory, the value of the work is lessened. The technical principles have been dealt with heretofore more thoroughly than the application of these principles, and it is the purpose of this book to consider more completely this practical aspect in the hope that a better understanding of the factors governing commercial practice may suggest means of improvement in shop methods.

It has been the author's attempt to make the chapters dealing with the heating problem illustrate the necessity for considering all of the elements entering into the production of a uniformly heated product; of heat application rather than details of construction; of the importance of the human element and the relation it bears to the ultimate results; of encouraging a broad and common-sense view of the principles governing this important operation and which generally have been overlooked; and, finally, of viewing the heating problem as an engineering proposition, adapting each fuel to proper furnace design and operation to meet the requirements of the problem in hand, and by so doing aim for the adoption of the standard heating unit in terms of finished product—"the cost of a unit of quantity of given quality."

He has attempted to make as practical as possible those chapters relating to steel and the effect of heat upon steel. Theories have been advanced only so far as has been thought necessary for a clear under-

standing of principles. Wherever possible, illustrations in the form of photomicrographs and charts have been given. The data given under the various types of heat-treated steels have been checked as far as possible and every effort has been made to be correct.

To the many friends who have aided him in the preparation of this book the author would express his sincere appreciation. Effort has been made to give due credit for cuts and data at the proper place, and for such as may not have been made acknowledgment is hereby rendered.

DENISON K. BULLENS

PHILADELPHIA,

October 1, 1915.

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TABLE I

TEMPERATURE CONVERSION TABLE

By Dr. Leonard Waldo

Reprint from *Metallurgical and Chemical Engineering*

C°	0	10	20	30	40	50	60	70	80	90		
-200	-328	-346	-364	-382	-400	-418	-436	-454	-492	-310		
-100	-148	-166	-184	-202	-220	-238	-256	-274	-292	-130		
-0	+32	+14	-4	-22	-40	-58	-76	-94	-112	-130		
0	32	50	68	86	104	122	140	158	176	194	C°	F°
100	212	230	248	266	284	302	320	338	356	374	1	1.8
200	392	410	428	446	464	482	500	518	536	554	2	3.6
300	572	590	608	626	644	662	680	698	716	734	3	5.4
400	752	770	788	806	824	842	860	878	892	914	4	7.2
500	932	950	968	986	1004	1022	1040	1058	1076	1094	5	9.0
600	1112	1130	1148	1166	1184	1202	1220	1238	1256	1274	6	10.8
700	1292	1310	1328	1346	1364	1382	1400	1418	1436	1454	7	12.6
800	1472	1490	1508	1526	1544	1562	1580	1598	1616	1634	8	14.4
900	1652	1670	1688	1706	1724	1742	1760	1778	1796	1814	9	16.2
1000	1832	1850	1868	1886	1904	1922	1940	1958	1976	1994	10	18.0
1100	2012	2030	2048	2066	2084	2102	2120	2138	2156	2174		
1200	2192	2210	2228	2246	2264	2282	2300	2318	2336	2354		
1300	2372	2390	2408	2426	2444	2462	2480	2498	2516	2534		
1400	2552	2570	2588	2606	2624	2642	2660	2678	2696	2714		
1500	2732	2750	2768	2786	2804	2822	2840	2858	2876	2894		
1600	2912	2930	2948	2966	2984	3002	3020	3038	3056	3074	F°	C°
1700	3092	3110	3128	3146	3164	3182	3200	3218	3236	3254	1	.56
1800	3272	3290	3308	3326	3344	3362	3380	3398	3416	3234	2	1.11
1900	3452	3470	3488	3506	3524	3542	3560	3578	3596	3614	3	1.67
2000	3632	3650	3668	3686	3704	3722	3740	3758	3776	3794	4	2.22
2100	3812	3830	3848	3866	3884	3902	3920	3938	3956	3974	5	2.78
2200	3992	4010	4028	4046	4064	4082	4100	4118	4136	4154	6	3.33
2300	4172	4190	4208	4226	4244	4262	4280	4298	4316	4334	7	3.89
2400	4352	4370	4388	4406	4424	4442	4460	4478	4496	4514	8	4.44
2500	4532	4550	4568	4586	4604	4622	4640	4658	4676	4694	9	5.00
2600	4712	4730	4748	4766	4784	4802	4820	4838	4856	4874	10	5.56
2700	4892	4910	4928	4946	4964	4982	5000	5018	5036	5054	11	6.11
2800	5072	5090	5108	5126	5144	5162	5180	5198	5216	5234	12	6.67
2900	5252	5270	5288	5306	5324	5342	5360	5378	5396	5414	13	7.22
3000	5432	5450	5468	5486	5504	5522	5540	5558	5576	5594	14	7.78
											15	8.33
3100	5612	5630	5648	5666	5684	5702	5720	5738	5756	5774	16	8.89
3200	5792	5810	5828	5846	5864	5882	5900	5918	5936	5954	17	9.44
3300	5972	5990	6008	6026	6044	6062	6080	6098	6116	6134	18	10.00
3400	6152	6170	6188	6206	6224	6242	6260	6278	6296	6314		
3500	6332	6350	6368	6386	6404	6422	6440	6458	6476	6494		
3600	6512	6530	6548	6566	6584	6602	6620	6638	6656	6674		
3700	6692	6710	6728	6746	6764	6782	6800	6818	6836	6854		
3800	6872	6890	6908	6926	6944	6962	6980	6998	7016	7034		
3900	7052	7070	7088	7106	7124	7142	7160	7178	7196	7214		
C°	0	10	20	30	40	50	60	70	80	90		

EXAMPLES: 1347°C 2444°F + 12°.6F = 2456°.6F: 3367°F = 1850°C + 2°.79C = 1852°.79C

# STEEL NUMBERING SYSTEMS

TABLE II

## BASIC OPEN-HEARTH AND ACID-BESSEMER CARBON STEELS

<i>AISI</i> Number	<i>Semi-</i> <i>finish</i>	<i>Bars</i>	<i>Wire</i> <i>Rods</i>	C	Mn	P (b)	S (b)	<i>1942 SAE</i> Number
C 1005	..	..	✓	0.06 max.	0.35 max.	0.04	0.05	....
C 1006 *	Q, R	✓	✓	0.08 max.	0.25-0.40	0.04	0.05	....
C 1008 *	Q, R	✓	✓	0.10 max.	0.30-0.50	0.04	0.05	1008
CB 1008	..	✓	✓	0.10 max.	.....	.....	.....	....
C 1009	Q	..	..	0.07-0.12	0.25-0.40	0.04	0.05	....
C 1010 *	Q, R	✓	✓	0.08-0.13	0.30-0.50	0.04	0.05	1010
C 1012	Q	✓	✓	0.10-0.15	0.30-0.50	0.04	0.05	....
CB 1012	..	✓	..	0.15 max.	.....	.....	.....	....
C 1013	..	..	✓	0.11-0.16	0.60-0.90	0.04	0.05	....
C 1014	Q	✓	✓	0.13-0.18	0.40-0.60	0.04	0.05	....
C 1015 *	Q, R	✓	✓	0.13-0.18	0.30-0.50	0.04	0.05	1015
C 1016	Q	✓	✓	0.13-0.18	0.60-0.90	0.04	0.05	{ X1015 1016
CB 1017	..	✓	..	0.10-0.25	.....	.....	.....	....
C 1017	Q	✓	✓	0.15-0.20	0.40-0.60	0.04	0.05	....
C 1018	Q	✓	✓	0.15-0.20	0.60-0.90	0.04	0.05	....
C 1019	Q	✓	✓	0.15-0.20	0.70-1.00	0.04	0.05	....
C 1020 *	Q, R	✓	✓	0.18-0.23	0.30-0.50	0.04	0.05	1020
C 1021	Q	✓	✓	0.18-0.23	0.40-0.60	0.04	0.05	....
C 1022	Q	✓	✓	0.18-0.23	0.70-1.00	0.04	0.05	{ X1020 1022
C 1023	Q	✓	✓	0.20-0.25	0.30-0.50	0.04	0.05	....
C 1024	..	✓	..	0.20-0.26	1.35-1.65	0.04	0.05	1024
C 1025 *	Q, R	✓	✓	0.22-0.28	0.30-0.50	0.04	0.05	1025
C 1026	Q	✓	✓	0.22-0.28	0.40-0.60	0.04	0.05	....
C 1027	..	..	✓	0.24-0.30	0.40-0.60	0.04	0.05	....
C 1029	Q	✓	..	0.25-0.31	0.60-0.90	0.04	0.05	....
C 1030	Q	✓	✓	0.28-0.34	0.60-0.90	0.04	0.05	1030
C 1031	Q	..	..	0.28-0.34	0.40-0.60	0.04	0.05	....
CB 1032	R	✓	..	0.25-0.40	.....	.....	.....	....
C 1033	Q	✓	..	0.30-0.36	0.60-0.90	0.04	0.05	....
C 1034	..	..	✓	0.32-0.38	0.50-0.70	0.04	0.05	....
D 1034	..	..	✓	0.32-0.38	0.50-0.70	0.05	0.05	....
C 1035	Q	✓	..	0.32-0.38	0.60-0.90	0.04	0.05	1035
C 1036	..	✓	..	0.32-0.39	1.20-1.50	0.04	0.05	1036
C 1037	C.r. strip	..	..	0.32-0.38	0.40-0.60	0.04	0.05	....
C 1038	Q	..	✓	0.35-0.42	0.60-0.90	0.04	0.05	....
C 1039	C.r. strip	..	..	0.37-0.44	0.40-0.60	0.04	0.05	....
C 1040	Q	✓	✓	0.37-0.44	0.60-0.90	0.04	0.05	1040
C 1041	..	..	✓	0.36-0.44	1.35-1.65	0.04	0.05	....
C 1042	Q	✓	..	0.40-0.47	0.60-0.90	0.04	0.05	....
C 1043	Q	✓	..	0.40-0.47	0.70-1.00	0.04	0.05	....
C 1044	..	..	✓	0.43-0.50	0.50-0.70	0.04	0.05	....
C 1045	Q	✓	..	0.43-0.50	0.60-0.90	0.04	0.05	1045
C 1046	Q	..	..	0.43-0.50	0.70-1.00	0.04	0.05	....
C 1047	C.r. strip	..	..	0.43-0.50	0.40-0.60	0.04	0.05	....



TABLE II (Continued)

## BASIC OPEN-HEARTH AND ACID-BESSEMER CARBON STEELS

AISI Number	Semi- finish	Bars	Wire Rods			P	S	1942 SAE Number
				C	Mn	(b)	(b)	
C 1049	C.r. strip	..	..	0.48-0.55	0.40-0.60	0.04	0.05	....
D 1049	..	..	✓	0.43-0.50	0.50-0.70	0.05	0.05	....
C 1050	Q	✓	..	0.48-0.55	0.60-0.90	0.04	0.05	1050
C 1051	..	..	✓	0.45-0.56	0.85-1.15	0.04	0.05	....
C 1052	..	✓	..	0.47-0.55	1.20-1.50	0.04	0.05	1052
C 1054	..	..	✓	0.50-0.60	0.50-0.70	0.04	0.05	....
D 1054	..	..	✓	0.50-0.60	0.50-0.70	0.05	0.05	....
C 1055	Q	✓	..	0.50-0.60	0.60-0.90	0.04	0.05	1055
C 1056	C.r. strip	..	..	0.50-0.60	0.40-0.60	0.04	0.05	....
C 1057	..	..	✓	0.50-0.61	0.85-1.15	0.04	0.05	....
C 1058	C.r. strip	..	..	0.55-0.65	0.40-0.60	0.04	0.05	....
C 1059	..	..	✓	0.55-0.65	0.50-0.70	0.04	0.05	....
D 1059	..	..	✓	0.55-0.65	0.50-0.70	0.05	0.05	....
C 1060	..	✓	✓	0.55-0.65	0.60-0.90	0.04	0.05	1060
C 1061	..	✓	..	0.54-0.65	0.75-1.05	0.04	0.05	....
C 1062	..	..	✓	0.54-0.65	0.85-1.15	0.04	0.05	....
C 1063	..	..	..	0.57-0.67	0.60-0.85	0.04	0.05	....
C 1064	..	✓	✓	0.60-0.70	0.50-0.70	0.04	0.05	....
D 1064	..	..	✓	0.60-0.70	0.50-0.70	0.05	0.05	....
C 1065	Strip	..	..	0.60-0.70	0.60-0.90	0.04	0.05	....
C 1066	..	✓	✓	0.60-0.71	0.80-1.10	0.04	0.05	{ X1065 1066
C 1068	..	✓	..	0.65-0.75	0.50 max.	0.04	0.05	....
C 1069	..	..	✓	0.65-0.75	0.50-0.70	0.04	0.05	....
D 1069	..	..	✓	0.65-0.75	0.40-0.60	0.05	0.05	....
C 1070	..	✓	..	0.65-0.75	0.70-1.00	0.04	0.05	1070
C 1074	..	✓	✓	0.70-0.80	0.50-0.70	0.04	0.05	....
D 1074	..	..	✓	0.70-0.80	0.40-0.60	0.05	0.05	....
C 1075	..	..	✓	0.70-0.80	0.60-0.80	0.04	0.05	....
C 1076	Q	..	..	0.65-0.85	0.60-0.85	0.04	0.05	....
C 1078	..	✓	✓	0.72-0.85	0.30-0.50	0.04	0.05	....
D 1078	..	..	✓	0.70-0.85	0.30-0.50	0.05	0.05	....
C 1080	..	✓	..	0.75-0.88	0.60-0.90	0.04	0.05	1080
D 1083	..	..	✓	0.80-0.95	0.30-0.50	0.05	0.05	....
C 1084	Q	..	..	0.80-0.93	0.60-0.90	0.04	0.05	....
C 1085	Q	✓	..	0.80-0.93	0.70-1.10	0.04	0.05	1085
C 1086	..	✓	✓	0.82-0.95	0.30-0.50	0.04	0.05	....
C 1090	Q	..	..	0.85-1.00	0.60-0.90	0.04	0.05	....
C 1095	..	✓	✓	0.90-1.05	0.30-0.50	0.04	0.05	1095
D 1095	..	..	✓	0.90-1.05	0.30-0.50	0.05	0.05	....
B 1006	..	..	✓	0.08 max.	0.45 max.	0.11 max.	0.06 max.	....
B 1008	R	✓	..	0.10 max.	0.30-0.50	0.11 max.	0.06 max.	....
B 1011	..	✓	✓	0.13 max.	0.50-0.70	0.11 max.	0.06 max.	....

\*Compositions are given for forging quality; rerolling quality differs slightly in analysis.

When silicon is specified in standard basic open-hearth steels, silicon may be ordered only as 0.10 per cent maximum; 0.10 to 0.20 per cent; or 0.15 to 0.30 per cent. In the case of many grades of basic open-hearth steel, special practice is necessary in order to comply with a specification including silicon.

Acid-Bessemer steel is not furnished with specified silicon content.

Code for prefixes:

A is basic open-hearth alloy steel.

B is acid-Bessemer carbon steel.

C is basic open-hearth carbon steel.

D is acid open-hearth carbon steel.

E is electric-furnace steel.

NE is National Emergency standard steel; designation promulgated by WPB.

Q is forging quality, or special-requirement quality.

R is rerolling quality billets.

TABLE III

FREE-CUTTING SULPHURIZED OR PHOSPHORIZED CARBON STEELS

AISI Number	Semi- finish	Bars	Wire Rods	C		Mn		P	S	1942 SAE Number
								(b)	(b)	
B 1106	..	..	✓	0.09 max.	0.50 max.	0.11 max.	0.04 -0.09	....	....	
C 1108	..	..	✓	0.08-0.13	0.50-0.70	0.045 max.	0.07 -0.12	....	....	
C 1109	Q	✓	✓	0.08-0.13	0.60-0.90	0.045 max.	0.08 -0.13	....	....	
B 1110	..	✓	✓	0.13 max.	0.60 max.	0.11 max.	0.045-0.075	....	....	
C 1110	..	✓	✓	0.08-0.13	0.60-0.90	0.045 max.	0.10 -0.15	....	....	
B 1111	..	✓	✓	0.08-0.13	0.60-0.90	0.09-0.13	0.10 -0.15	....	....	1111
C 1111	..	..	✓	0.08-0.13	0.60-0.90	0.045 max.	0.16 -0.23	....	....	
B 1112	..	✓	✓	0.08-0.13	0.60-0.90	0.09-0.13	0.16 -0.23	....	....	1112
C 1112	..	✓	..	0.10-0.16	1.00-1.30	0.045 max.	0.08 -0.13	....	....	
B 1113	..	✓	✓	0.08-0.13	0.60-0.90	0.09-0.13	0.24 -0.33	....	....	{ X1112 1113
C 1113	Q	✓	..	0.10-0.16	1.00-1.30	0.045 max.	0.24 -0.33	....	....	
C 1114	R	..	..	0.12-0.18	0.45-0.65	0.045 max.	0.075-0.15	....	....	
C 1115	Q	✓	✓	0.13-0.18	0.70-1.00	0.045 max.	0.10 -0.15	....	....	1115
C 1116	Q	✓	✓	0.14-0.20	1.10-1.40	0.045 max.	0.16 -0.23	....	....	
C 1117	..	✓	✓	0.14-0.20	1.00-1.30	0.045 max.	0.08 -0.13	....	....	{ X1314 1117
C 1118	Q	✓	✓	0.14-0.20	1.30-1.60	0.045 max.	0.08 -0.13	....	....	{ X1315 1118
C 1119	..	..	✓	0.14-0.20	1.35-1.65	0.045 max.	0.16 -0.23	....	....	
C 1120 *	Q, R	✓	✓	0.18-0.23	0.60-0.90	0.045 max.	0.08 -0.13	....	....	
C 1121	Q	✓	✓	0.18-0.23	0.70-1.00	0.045 max.	0.08 -0.13	....	....	
C 1122	Q	✓	✓	0.17-0.23	1.35-1.65	0.045 max.	0.08 -0.13	....	....	
C 1132	Q	✓	✓	0.27-0.34	1.35-1.65	0.045 max.	0.08 -0.13	....	....	{ X1330 1132
C 1137	Q	✓	✓	0.32-0.39	1.35-1.65	0.045 max.	0.08 -0.13	....	....	{ X1335 1137
C 1140 †	Q	..	..	0.37-0.44	0.60-0.90	0.045 max.	0.04 -0.07	....	....	
C 1141	..	✓	..	0.37-0.45	1.35-1.65	0.045 max.	0.08 -0.13	....	....	{ X1340 1141
C 1144	..	✓	..	0.40-0.48	1.35-1.65	0.045 max.	0.24 -0.33	....	....	
C 1145 †	Q	..	..	0.42-0.49	0.70-1.00	0.045 max.	0.04 -0.07	....	....	1145
C 1205	Q	..	✓	0.08 max.	0.25-0.40	0.04-0.07	0.05 max.	....	....	
C 1206	R	..	✓	0.08 max.	0.25-0.40	0.06-0.10	0.05 max.	....	....	
C 1209	R	..	✓	0.08-0.13	0.30-0.50	0.04-0.07	0.05 max.	....	....	
C 1210	Q	..	✓	0.08-0.13	0.30-0.50	0.06-0.10	0.05 max.	....	....	
C 1217	..	✓	..	0.14-0.19	0.70-1.00	0.09-0.13	0.20-0.29	....	....	

\* Compositions given are for forging quality; reolling quality differs slightly in analysis.

(b) Phosphorus and sulphur are maximum for basic open-hearth and acid-Bessemer steels; sulphurized or phosphorized steels are not subject to check analysis for S or P.

Acid-Bessemer steels (B series) are not furnished with specified silicon content.

† Standard steels C 1140 and C 1145 may be ordered with silicon content either as 0.10% max., 0.10 to 0.20%, or 0.15 to 0.30%.

TABLE IV

OPEN-HEARTH ALLOY AND ELECTRIC-FURNACE CARBON AND ALLOY STEELS

Blooms, Billets, Slabs, Bars, and Hot-Rolled Strip

AISI Number	C	Mn	P		S Si *	Ni	Cr	Mo	'42 SAE Number
			Max.*	Max.*					
A 1330	0.28-0.33	1.60-1.90	0.040	0.040	0.20-0.35	.....	.....	.....	1330
A 1335	0.33-0.38	1.60-1.90	0.040	0.040	0.20-0.35	.....	.....	.....	1335
A 1340	0.38-0.43	1.60-1.90	0.040	0.040	0.20-0.35	.....	.....	.....	1340
A 1345	0.43-0.48	1.60-1.90	0.040	0.040	0.20-0.35	.....	.....	.....	....
A 1350	0.48-0.53	1.60-1.90	0.040	0.040	0.20-0.35	.....	.....	.....	....
A 2317	0.15-0.20	0.40-0.60	0.040	0.040	0.20-0.35	3.25-3.75	.....	.....	{ 2315 2317
A 2330	0.28-0.33	0.60-0.80	0.040	0.040	0.20-0.35	3.25-3.75	.....	.....	2330
A 2335	0.33-0.38	0.60-0.80	0.040	0.040	0.20-0.35	3.25-3.75	.....	.....	....
A 2340	0.38-0.43	0.70-0.90	0.040	0.040	0.20-0.35	3.25-3.75	.....	.....	2340
A 2345	0.43-0.48	0.70-0.90	0.040	0.040	0.20-0.35	3.25-3.75	.....	.....	2345
E 2512	0.09-0.14	0.45-0.60	0.025	0.025	0.20-0.35	4.75-5.25	.....	.....	....
A 2515	0.12-0.17	0.40-0.60	0.040	0.040	0.20-0.35	4.75-5.25	.....	.....	2515
E 2517	0.15-0.20	0.45-0.60	0.025	0.025	0.20-0.35	4.75-5.25	.....	.....	....
A 3115	0.13-0.18	0.40-0.60	0.040	0.040	0.20-0.35	1.10-1.40	0.55-0.75	.....	3115
A 3120	0.17-0.22	0.60-0.80	0.040	0.040	0.20-0.35	1.10-1.40	0.55-0.75	.....	3120
A 3130	0.28-0.33	0.60-0.80	0.040	0.040	0.20-0.35	1.10-1.40	0.55-0.75	.....	3130
A 3135	0.33-0.38	0.60-0.80	0.040	0.040	0.20-0.35	1.10-1.40	0.55-0.75	.....	3135
A 3140	0.38-0.43	0.70-0.90	0.040	0.040	0.20-0.35	1.10-1.40	0.55-0.75	.....	3140
A 3141	0.38-0.43	0.70-0.90	0.040	0.040	0.20-0.35	1.10-1.40	0.70-0.90	.....	{ X3140 3141
A 3145	0.43-0.48	0.70-0.90	0.040	0.040	0.20-0.35	1.10-1.40	0.70-0.90	.....	3145
A 3150	0.48-0.53	0.70-0.90	0.040	0.040	0.20-0.35	1.10-1.40	0.70-0.90	.....	3150
A 3240	0.38-0.45	0.40-0.60	0.040	0.040	0.20-0.35	1.65-2.00	0.90-1.20	.....	3240
E 3310	0.08-0.13	0.45-0.60	0.025	0.025	0.20-0.35	3.25-3.75	1.40-1.75	.....	{ 3310 3312
E 3316	0.14-0.19	0.45-0.60	0.025	0.025	0.20-0.35	3.25-3.75	1.40-1.75	.....	....
A 4023	0.20-0.25	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4023
A 4024	0.20-0.25	0.70-0.90	0.040	(a)	0.20-0.35	.....	0.20-0.30	.....	....
A 4027	0.25-0.30	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4027
A 4028	0.25-0.30	0.70-0.90	0.040	(a)	0.20-0.35	.....	0.20-0.30	.....	....
A 4032	0.30-0.35	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4032
A 4037	0.35-0.40	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4037
A 4042	0.40-0.45	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4042
A 4047	0.45-0.50	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4047
A 4063	0.60-0.67	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4063
A 4068	0.64-0.72	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.20-0.30	.....	4068
A 4119	0.17-0.22	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.40-0.60	0.20-0.30	4119
A 4120	0.17-0.22	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.60-0.80	0.20-0.30	....
A 4125	0.23-0.28	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.40-0.60	0.20-0.30	4125
A 4130	0.28-0.33	0.40-0.60	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	{ X4130 4130
A 4131	0.28-0.33	0.50-0.70	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.14-0.19	....
E 4132	0.30-0.35	0.40-0.60	0.025	0.025	0.20-0.35	.....	0.80-1.10	0.18-0.25	....
A 4134	0.32-0.37	0.40-0.60	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	....
E 4135	0.33-0.38	0.70-0.90	0.025	0.025	0.20-0.35	.....	0.80-1.10	0.18-0.25	....
A 4137	0.35-0.40	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	4137
E 4137	0.35-0.40	0.70-0.90	0.025	0.025	0.20-0.35	.....	0.80-1.10	0.18-0.25	....
A 4140	0.38-0.43	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	4140
A 4141	0.38-0.43	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.14-0.19	....
A 4142	0.40-0.45	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	....
A 4143	0.40-0.45	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.30-0.40	....
A 4145	0.43-0.48	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	4145
A 4147	0.45-0.52	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	....

TABLE IV (Continued)

OPEN-HEARTH ALLOY AND ELECTRIC-FURNACE CARBON AND ALLOY STEELS

Blooms, Billets, Slabs, Bars, and Hot-Rolled Strip

AISI Number	P					S			Mo	'42 SAE Number
	C	Mn	Max.*	Max.*	Si *	Ni	Cr			
A 4150	0.46-0.53	0.75-1.00	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15-0.25	4150	
E 4150	0.48-0.53	0.70-0.90	0.025	0.025	0.20-0.35	.....	0.80-1.10	0.20-0.27	....	
A 4317	0.15-0.20	0.45-0.65	0.040	0.040	0.20-0.35	1.65-2.00	0.40-0.60	0.20-0.30	....	
A 4320	0.17-0.22	0.45-0.65	0.040	0.040	0.20-0.35	1.65-2.00	0.40-0.60	0.20-0.30	4320	
A 4337	0.35-0.40	0.60-0.80	0.040	0.040	0.20-0.35	1.65-2.00	0.60-0.80	0.30-0.40	....	
E 4337	0.35-0.40	0.60-0.80	0.025	0.025	0.20-0.35	1.65-2.00	0.70-0.90	0.23-0.30	....	
A 4340	0.38-0.43	0.60-0.80	0.040	0.040	0.20-0.35	1.65-2.00	0.70-0.90	0.20-0.30	X4340 4340	
E 4342	0.40-0.45	0.60-0.80	0.025	0.025	0.20-0.35	1.65-2.00	0.70-0.90	0.23-0.30		....
A 4608	0.06-0.11	0.40 max.	0.040	0.040	0.25 max.	1.40-1.75	.....	0.15-0.25	....	
A 4615	0.13-0.18	0.45-0.65	0.040	0.040	0.20-0.35	1.65-2.00	.....	0.20-0.30	4615	
E 4617	0.15-0.20	0.45-0.65	0.025	0.025	0.20-0.35	1.65-2.00	.....	0.20-0.27	....	
A 4620	0.17-0.22	0.45-0.65	0.040	0.040	0.20-0.35	1.65-2.00	.....	0.20-0.30	4620	
E 4620	0.17-0.22	0.45-0.60	0.025	0.025	0.20-0.35	1.65-2.00	.....	0.20-0.27	....	
A 4621	0.18-0.23	0.70-0.90	0.040	0.040	0.20-0.35	1.65-2.00	.....	0.20-0.30	....	
A 4640	0.38-0.43	0.60-0.80	0.040	0.040	0.20-0.35	1.65-2.00	.....	0.20-0.30	4640	
E 4640	0.38-0.43	0.60-0.80	0.025	0.025	0.20-0.35	1.65-2.00	.....	0.20-0.27	....	
A 4645	0.43-0.48	0.60-0.80	0.040	0.040	0.20-0.35	1.65-2.00	.....	0.20-0.30	....	
A 4815	0.13-0.18	0.40-0.60	0.040	0.040	0.20-0.35	3.25-3.75	.....	0.20-0.30	4815	
A 4820	0.18-0.23	0.50-0.70	0.040	0.040	0.20-0.35	3.25-3.75	.....	0.20-0.30	4820	
A 5120	0.17-0.22	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.70-0.90	.....	5120	
A 5130	0.28-0.33	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.80-1.10	.....	....	
A 5140	0.38-0.43	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.70-0.90	.....	5140	
A 5145	0.43-0.48	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.70-0.90	.....	....	
A 5150	0.48-0.55	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.70-0.90	.....	5150	
A 5152	0.45-0.55	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.90-1.20	.....	....	
E 52095	0.95-1.10	0.25-0.45	0.025	0.025	0.20-0.35	.....	0.40-0.60	.....	....	
E 52098	0.95-1.10	0.25-0.45	0.025	0.025	0.20-0.35	.....	0.90-1.15	.....	....	
E 52100(b)	0.95-1.10	0.25-0.45	0.025	0.025	0.20-0.35	.....	1.20-1.50	.....	....	
E 52101	0.95-1.10	0.25-0.45	0.025	0.025	0.20-0.35	.....	1.30-1.60	.....	....	
A 6120	0.17-0.22	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.70-0.90	0.10 min.V	....	
A 6145	0.43-0.48	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15 min.V	....	
SAE 6150	0.48-0.55	0.65-0.90	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.15 min.V	6150	
E 6151	0.47-0.53	0.70-0.90	0.025	0.025	0.20-0.35	.....	0.80-1.10	0.15 min.V	....	
A 6152	0.48-0.55	0.70-0.90	0.040	0.040	0.20-0.35	.....	0.80-1.10	0.10 min.V	....	
A 9255	0.50-0.60	0.70-0.95	0.040	0.040	1.80-2.20	.....	.....	.....	....	
A 9280(c)	0.55-0.65	0.75-1.00	0.040	0.040	1.80-2.20	.....	.....	.....	....	
A 9282	0.55-0.65	0.75-1.00	0.040	0.040	1.80-2.20	.....	0.20-0.40	.....	....	

\* Lowest standard maximum phosphorus or sulphur content for acid open-hearth or acid electric-furnace alloy steel is 0.05% each; silicon is 0.15% min.

(a) Sulphur range: 0.035 to 0.050%.

(b) See 52100 variations, page xxvii.

(c) See 9261 variation, page xxvii.

The 8600 and 8700 series, originally designated as N.E. steels, and given on the next page, were later incorporated into the regular series.

TABLE V

## THE NATIONAL EMERGENCY STEELS

Revised May 26, 1944

In this revision of the N.E. steel composition limits, the 8700 series has been restored in a wider variety of carbon ranges. A new series, the 9700 series, has been established having hardenability characteristics similar to those of the 4000 series. A new series, the 9800 series, has been established having hardenability characteristics about half-way between those of the 8700 series and those of the 4300 series, in the thorough-hardening carbon ranges. A new series, the 9900 series, has been established in the low-carbon ranges, designed especially for carburizing and having hardenability characteristics similar to or slightly in excess of the 8700 series.

Classi- fication	Designation	Composition					
		C	Mn	Si	Ni	Cr	Mo
C-Mn	N.E. 1330	0.28-0.33	1.60-1.90	0.20-0.35	.....	.....	.....
	N.E. 1335	0.33-0.38	1.60-1.90	0.20-0.35	.....	.....	.....
	N.E. 1340	0.38-0.43	1.60-1.90	0.20-0.35	.....	.....	.....
	N.E. 1345	0.43-0.48	1.60-1.90	0.20-0.35	.....	.....	.....
	N.E. 1350	0.48-0.53	1.60-1.90	0.20-0.35	.....	.....	.....
Ni-Cr-Mo 8000 and 8400 series, see below	N.E. 8612 *	0.10-0.15	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8615	0.13-0.18	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8617	0.15-0.20	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8620	0.18-0.23	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8622	0.20-0.25	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8625	0.23-0.28	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8627	0.25-0.30	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8630	0.28-0.33	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8632	0.30-0.35	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8635	0.33-0.38	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8637	0.35-0.40	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8640	0.38-0.43	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8642	0.40-0.45	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8645	0.43-0.48	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8647	0.45-0.50	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8650	0.48-0.53	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
	N.E. 8712	0.10-0.15	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8715	0.13-0.18	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8717	0.15-0.20	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8720	0.18-0.23	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8722	0.20-0.25	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8725	0.23-0.28	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8727	0.25-0.30	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8730	0.28-0.33	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8732	0.30-0.35	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8735	0.33-0.38	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8737	0.35-0.40	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8740	0.38-0.43	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8742	0.40-0.45	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8745	0.43-0.48	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8747	0.45-0.50	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 8750	0.48-0.53	0.75-1.00	0.20-0.35	0.40-0.70	0.40-0.60	0.20-0.30
	N.E. 9415	0.13-0.18	0.80-1.10	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9417	0.15-0.20	0.80-1.10	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9420	0.18-0.23	0.80-1.10	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9422	0.20-0.25	0.80-1.10	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9425	0.23-0.28	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9427	0.25-0.30	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9430	0.28-0.33	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9432	0.30-0.35	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9435	0.33-0.38	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9437	0.35-0.40	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9440	0.38-0.43	0.90-1.20	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9442	0.40-0.45	1.00-1.30	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9445	0.43-0.48	1.00-1.30	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9447	0.45-0.50	1.20-1.50	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15
	N.E. 9450	0.48-0.53	1.20-1.50	0.20-0.35	0.30-0.60	0.30-0.50	0.08-0.15

TABLE V (Continued)

THE NATIONAL EMERGENCY STEELS

Revised May 26, 1944

Classification	Designation	Composition					
		C	Mn	Si	Ni	Cr	Mo
	9500 and 9600 series, see below						
	N.E. 9722	0.20-0.25	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9727	0.25-0.30	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9732	0.30-0.35	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9737	0.35-0.40	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9742	0.40-0.45	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9745	0.43-0.48	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9747	0.45-0.50	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9750	0.48-0.53	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9763	0.60-0.67	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9768	0.64-0.72	0.50-0.80	0.20-0.35	0.40-0.70	0.10-0.25	0.15-0.25
	N.E. 9830	0.28-0.33	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9832	0.30-0.35	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9835	0.33-0.38	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9837	0.35-0.40	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9840	0.38-0.43	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9842	0.40-0.45	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9845	0.43-0.48	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9847	0.45-0.50	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9850	0.48-0.53	0.70-0.90	0.20-0.35	0.85-1.15	0.70-0.90	0.20-0.30
	N.E. 9912	0.10-0.15	0.50-0.70	0.20-0.35	1.00-1.30	0.40-0.60	0.20-0.30
	N.E. 9915	0.13-0.18	0.50-0.70	0.20-0.35	1.00-1.30	0.40-0.60	0.20-0.30
	N.E. 9917	0.15-0.20	0.50-0.70	0.20-0.35	1.00-1.30	0.40-0.60	0.20-0.30
	N.E. 9920	0.18-0.23	0.50-0.70	0.20-0.35	1.00-1.30	0.40-0.60	0.20-0.30
	N.E. 9922	0.20-0.25	0.50-0.70	0.20-0.35	1.00-1.30	0.40-0.60	0.20-0.30
	N.E. 9925	0.23-0.28	0.50-0.70	0.20-0.35	1.00-1.30	0.40-0.60	0.20-0.30
Si-Mn	N.E. 9261	0.55-0.65	0.70-1.00	1.80-2.20	.....	0.10-0.25	.....
C-Cr	N.E. 52100A	0.95-1.10	0.25-0.45	0.20-0.35	0.35 max.	1.30-1.60	0.08 max.
	N.E. 52100B	0.95-1.10	0.25-0.45	0.20-0.35	0.35 max.	0.90-1.15	0.08 max.
	N.E. 52100C	0.95-1.10	0.25-0.45	0.20-0.35	0.35 max.	0.40-0.60	0.08 max.

Other N.E. compositions are, or were, as follows: S and P each 0.04 max.

N.E. 8020	0.18-0.23	1.00-1.30	0.20-0.35	.....	.....	0.10-0.20
N.E. 8024	0.22-0.28	1.00-1.30	0.20-0.35	.....	.....	0.10-0.20
N.E. 8124	0.22-0.28	1.30-1.60	0.20-0.35	.....	.....	0.25-0.35
N.E. 8253	0.30-0.36	1.30-1.60	0.20-0.35	.....	.....	0.10-0.20
N.E. 8245	0.42-0.49	1.30-1.60	0.20-0.35	.....	.....	0.10-0.20
N.E. 8359	0.35-0.42	1.30-1.60	0.20-0.35	.....	.....	0.20-0.30
N.E. 8442 *	0.40-0.45	1.30-1.60	0.20-0.35	.....	.....	0.30-0.40
N.E. 8447	0.43-0.50	1.30-1.60	0.20-0.35	.....	.....	0.30-0.40
N.E. 8547	0.43-0.50	1.30-1.60	0.20-0.35	.....	.....	0.40-0.60
N.E. 8613	0.12-0.17	0.70-0.90	0.20-0.35	0.40-0.70	0.40-0.60	0.15-0.25
N.E. 8724	0.22-0.28	0.70-0.95	0.20-0.35	0.40-0.60	0.40-0.60	0.20-0.30
N.E. 8739	0.35-0.42	0.75-1.00	0.20-0.35	0.40-0.60	0.40-0.60	0.20-0.30
N.E. 8744	0.40-0.47	0.75-1.00	0.20-0.35	0.40-0.60	0.40-0.60	0.20-0.30
N.E. 8749	0.45-0.52	0.75-1.00	0.20-0.35	0.40-0.60	0.40-0.60	0.20-0.30
N.E. 8817	0.15-0.20	0.70-0.95	0.20-0.35	0.40-0.60	0.40-0.60	0.30-0.40
N.E. 8949	0.45-0.52	1.00-1.35	0.20-0.35	0.40-0.60	0.40-0.60	0.30-0.40
N.E. 9537 *	0.35-0.40	1.20-1.50	0.40-0.60	0.40-0.70	0.40-0.60	0.15-0.25
N.E. 9540 *	0.38-0.43	1.20-1.50	0.40-0.60	0.40-0.70	0.40-0.60	0.15-0.25
N.E. 9542 *	0.40-0.45	1.20-1.50	0.40-0.60	0.40-0.70	0.40-0.60	0.15-0.25
N.E. 9550 *	0.48-0.53	1.20-1.50	0.40-0.60	0.40-0.70	0.40-0.60	0.15-0.25
N.E. 9630	0.28-0.33	1.20-1.50	0.40-0.60	0.40-0.70	0.40-0.60	0.15-0.25
N.E. 9635	0.33-0.38	1.20-1.50	0.40-0.60	.....	0.40-0.60	.....
N.E. 9687	0.35-0.40	1.20-1.50	0.40-0.60	.....	0.40-0.60	.....
N.E. 9640	0.38-0.43	1.20-1.50	0.40-0.60	.....	0.40-0.60	.....
N.E. 9642	0.40-0.45	1.30-1.60	0.40-0.60	.....	0.40-0.60	.....
N.E. 9645	0.43-0.48	1.30-1.60	0.40-0.60	.....	0.40-0.60	.....
N.E. 9650	0.48-0.53	1.30-1.60	0.40-0.60	.....	0.40-0.60	.....

\* Recommended for large sections only.

Many of the N.E. compositions are no longer available, but technical literature often refers to such steels, requiring that even those now considered obsolete be recorded.

TABLE VI

## AMERICAN IRON AND STEEL INSTITUTE LIST OF H STEELS; CHEMICAL COMPOSITION RANGES

## Electric-Furnace or Open-Hearth Bars, Billets, or Blooms

SAE or AISI Designation	Chemical Composition *					
	C	Mn	Si	Ni	Cr	Mo
2512H	0.08-0.15	0.35-0.65	0.20-0.35	4.70-5.30	.....	.....
2515H	0.11-0.18	0.35-0.65	0.20-0.35	4.70-5.30	.....	.....
2517H	0.16-0.21	0.35-0.65	0.20-0.35	4.70-5.30	.....	.....
3310H	0.07-0.14	0.35-0.65	0.20-0.35	3.20-3.80	1.35-1.75	.....
3316H	0.13-0.20	0.35-0.65	0.20-0.35	3.20-3.80	1.35-1.75	.....
4130H	0.27-0.34	0.35-0.65	0.20-0.35	.....	0.80-1.15	0.15-0.25
4132H	0.30-0.37	0.35-0.65	0.20-0.35	.....	0.80-1.15	0.15-0.25
4135H	0.32-0.39	0.60-0.95	0.20-0.35	.....	0.80-1.15	0.15-0.25
4137H	0.35-0.43	0.60-0.95	0.20-0.35	.....	0.80-1.15	0.15-0.25
4140H	0.37-0.45	0.70-1.05	0.20-0.35	.....	0.80-1.15	0.15-0.25
4142H	0.40-0.48	0.70-1.05	0.20-0.35	.....	0.80-1.15	0.15-0.25
4145H	0.42-0.50	0.70-1.05	0.20-0.35	.....	0.80-1.15	0.15-0.25
4147H	0.44-0.52	0.70-1.05	0.20-0.35	.....	0.80-1.15	0.15-0.25
4150H	0.46-0.54	0.70-1.05	0.20-0.35	.....	0.80-1.15	0.15-0.25
4317H	0.14-0.21	0.40-0.70	0.20-0.35	1.50-2.00	0.35-0.65	0.20-0.30
4320H	0.16-0.23	0.40-0.70	0.20-0.35	1.50-2.00	0.35-0.65	0.20-0.30
4340H	0.37-0.45	0.60-0.95	0.20-0.35	1.50-2.00	0.65-0.95	0.20-0.30
4620H	0.17-0.24	0.40-0.70	0.20-0.35	1.50-2.00	.....	0.20-0.30
4640H	0.37-0.45	0.55-0.85	0.20-0.35	1.50-2.00	.....	0.20-0.30
4815H	0.12-0.19	0.35-0.65	0.20-0.35	3.20-3.80	.....	0.20-0.30
4820H	0.17-0.24	0.45-0.75	0.20-0.35	3.20-3.80	.....	0.20-0.30
8620H	0.17-0.24	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8622H	0.20-0.27	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8625H	0.22-0.29	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8627H	0.25-0.32	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8630H	0.27-0.34	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8632H	0.30-0.37	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8635H	0.32-0.39	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8637H	0.35-0.43	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8640H	0.37-0.45	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8642H	0.40-0.48	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8645H	0.42-0.50	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8647H	0.44-0.52	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8650H	0.46-0.54	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8720H	0.17-0.24	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8722H	0.20-0.27	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8725H	0.22-0.29	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8727H	0.25-0.32	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8730H	0.27-0.34	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8732H	0.30-0.37	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8735H	0.32-0.39	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8737H	0.35-0.43	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8740H	0.37-0.45	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30

TABLE VI (Continued)

## AMERICAN IRON AND STEEL INSTITUTE LIST OF H STEELS; CHEMICAL COMPOSITION RANGES

## Electric-Furnace or Open-Hearth Bars, Billets, or Blooms

SAE or AISI Designation	Chemical Composition *					
	C	Mn	Si	Ni	Cr	Mo
8742H	0.40-0.48	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8745H	0.42-0.50	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8747H	0.44-0.52	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
8750H	0.46-0.54	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.20-0.30
9420H	0.17-0.24	0.80-1.15	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9422H	0.20-0.27	0.80-1.15	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9425H	0.22-0.29	0.80-1.15	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9427H	0.25-0.32	0.80-1.15	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9430H	0.27-0.34	0.85-1.25	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9432H	0.30-0.37	0.85-1.25	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9435H	0.32-0.39	0.85-1.25	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9437H	0.35-0.43	0.85-1.25	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9440H	0.37-0.45	0.85-1.25	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9442H	0.40-0.48	0.95-1.35	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9445H	0.42-0.50	0.95-1.35	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9447H	0.44-0.52	1.15-1.55	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15
9450H	0.46-0.54	1.15-1.55	0.20-0.35	0.25-0.65	0.25-0.55	0.08-0.15

These chemical compositions apply to steels produced to definite hardenability limits. Composition limits and hardenability bands are intended to apply to steels manufactured to "fine-grained steel practice."

*Sizes and Shapes.* The ranges and limits in this table apply only to material not exceeding 100 sq in. in cross-sectional area, or 18 in. in width, or 7000 lb in weight, per piece as the total product of the ingot, and exclude all plate, shapes, sheet, strip and slabs.

\* *Permissible Chemical Variations.* Phosphorus and sulphur in open-hearth steel to be 0.040% max. each.

Phosphorus and sulphur in electric-furnace steel to be 0.025% max. each.

Small quantities of certain elements may be found in alloy steel which are not specified or required. These elements are to be considered as incidental and acceptable to the following maximum amounts: Copper 0.35%, nickel 0.25%, chromium 0.20%, molybdenum 0.06%.

The chemical ranges and limits shown are subject to the standard permissible variations for check analysis over the maximum limit or under the minimum limit. These permissible variations are: 0.01 for all ranges of carbon; 0.03 for manganese up to and including 0.90%, and 0.04 for over 0.90 to 1.55% inclusive; 0.005 for either phosphorus or sulphur; 0.02 for silicon; 0.03 for nickel up to and including 1.00%, 0.05 for nickel over 1.00 up to and including 2.00%, and 0.07 over 2.00 up to and including 5.30%; 0.03 for chromium up to 0.90% inclusive, and 0.05 for chromium over 0.90 to 1.75% inclusive; 0.01 for molybdenum up to 0.20% inclusive, 0.02 for molybdenum over 0.20 to 0.30% inclusive.

*Quality Conditions.* All conditions and quality features, except as detailed above, shall be in accordance with the regulation, shown in the American Iron and Steel Institute's *Steel Products Manual*, Section 10 on Alloy Steels.



TABLE VI (Continued)

## LATER H STEELS

In June 1947 the list of H steels was expanded to include the following:

SAE or AISI Designation	Chemical Composition					
	C	Mn	Si	Ni	Cr	Mo
1320H	0.17-0.24	1.50-2.00	0.20-0.35	.....	.....	.....
1330H	0.27-0.34	1.50-2.00	0.20-0.35	.....	.....	.....
1335H	0.32-0.39	1.50-2.00	0.20-0.35	.....	.....	.....
1340H	0.37-0.45	1.50-2.00	0.20-0.35	.....	.....	.....
4812H	0.10-0.17	0.30-0.60	0.20-0.35	3.20-3.80	.....	0.20-0.30
4817H	0.14-0.21	0.35-0.65	0.20-0.35	3.20-3.80	.....	0.20-0.30
5140H	0.37-0.45	0.60-0.95	0.20-0.35	.....	0.65-0.95	.....
5145H	0.42-0.50	0.60-0.95	0.20-0.35	.....	0.65-0.95	.....
5150H	0.46-0.54	0.60-0.95	0.20-0.35	.....	0.65-0.95	.....
6150H	0.46-0.54	0.60-0.95	0.20-0.35	.....	0.80-1.15	..... V, 0.15 min.
8617H	0.14-0.21	0.60-0.95	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8641H	0.37-0.45	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25 S, 0.04-0.06
8655H	0.50-0.60	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
8660H	0.55-0.65	0.70-1.05	0.20-0.35	0.35-0.75	0.35-0.65	0.15-0.25
9260H	0.55-0.65	0.70-1.05	1.80-2.20	.....	.....	.....
9261H	0.55-0.65	0.70-1.05	1.80-2.20	.....	0.05-0.35	.....
9262H	0.55-0.65	0.70-1.05	1.80-2.20	.....	0.20-0.50	.....

The following were omitted:

8722H  
8725H  
8227H  
8332H  
9420H  
9422H  
9425H  
9427H  
9430H  
9435H  
9447H  
9450H

# METALLURGICAL JOURNALS

**Specifications, Cross Indices.** A variety of specifications exist for closely similar steels. Useful lists of these are:

Cross index of chemical equivalent specifications and identification code, 1945, General Motors Corp., Detroit, Mich.

Preferred aeronautical steel specifications, *Steel*, V. 114, March 13, 1944, pp. 104-08.

**Abbreviations.** In the bibliographies at the ends of the chapters, the names of journals cited are abbreviated, as follows:

## Journals of National Societies

ASM—American Society for Metals, 7301 Euclid Avenue, Cleveland, Ohio: *Metals Handbook* (periodically revised); *Metal Progress* (monthly); *The Metals Review* (monthly—current abstracts of current literature); *ASM Metal Literature Review* (annual). Many special volumes of symposia or of special lecture series are issued in book form.

ASST—American Society for Steel Treating, 7301 Euclid Avenue, Cleveland, Ohio (earlier name of ASM).

AIME—American Institute of Mining & Metallurgical Engineers, 29 West 39th Street, New York 18, N. Y.: Some articles on heat treatment are in the annual *Transactions* of the Iron and Steel Division. Others appear as "Technical Publications" in *Metals Technology* (monthly).

ASTM—American Society for Testing Materials, 1916 Race Street, Philadelphia 3, Pa.: *Proceedings* (annual); *Bulletin* (monthly); Standard and Tentative Specifications, both for materials and methods of testing, are published every few years. Those for metals are in a separate volume.

AFA—American Foundrymen's Association, 222 West Adams Street, Chicago, Ill.: *Transactions* (annual); *Bulletin* (monthly); *Handbooks* (periodically revised).

SAE—Society of Automotive Engineers, 29 West 39th Street, New York 18, N. Y.: *Transactions* (annual); *Journal* (monthly).

AISI—American Iron and Steel Institute, 350 Fifth Avenue, New York, N. Y.: *Yearbook*.

I.&S.I.—(British) Iron & Steel Institute, 4 Grosvenor Gardens, London SW 1, England: *Proceedings and Abstracts* (semiannual); *Current Abstracts* (monthly).

AISI—American Iron & Steel Institute, 4 Grosvenor Gardens, London SW 1, England: *Proceedings and Abstracts* (semiannual); *Current Abstracts* (monthly).

ASME—American Society of Mechanical Engineers, 29 West 39th Street, New York 18, N. Y.: *Transactions* (annual); *Mechanical Engineering* (monthly); *Journal of Applied Mechanics* (monthly).

AWS—American Welding Society, 29 West 39th Street, New York 18, N. Y.:  
*Welding Journal* (monthly); Welding Research Council, separate (monthly);  
*Transactions* (annual).

**Other Metallurgical Journals Often Cited**

*Iron Age*—100 East 42nd Street, New York 17, N. Y. (weekly).

*Steel*—Penton Building, Cleveland 13, Ohio (weekly).

*The Foundry*—Penton Building, Cleveland 13, Ohio (monthly).

*Machine Design*—Penton Building, Cleveland 13, Ohio (monthly).

*Materials and Methods* (formerly *Metals and Alloys*)—330 West 42nd Street,  
New York 18, N. Y. (monthly).

*Product Engineering*—330 West 42nd Street, New York 18, N. Y. (monthly).

*Canadian Metals and Metallurgical Industries*—137 Wellington Street, Toronto  
1, Ontario, Canada.

*Industrial Heating*—Union Trust Building, Pittsburgh, Pa.

*Iron and Steel*—Dorset House, Stamford Street, London SE 1, England.

*Journal of Research*—National Bureau of Standards, Government Printing  
Office, Washington, D. C. (Earlier, Scientific Papers of the Bureau of  
Standards, Technologic Papers of the Bureau of Standards).

*Metal Treatment*—49 Wellington Street, Strand, London WC 2, England.

*Metallurgia*—21 Albion Street, Gaythorn, Manchester 1, England.

*Sheet Metal Industries*—49 Wellington Street, Strand, London WC 2, England.

*Steel Processing*—108 Southfield Street, Pittsburgh 30, Pa.

*Wire and Wire Products*—300 Main Street, Stamford, Conn.

## SECTION I. TERMINOLOGY

### CHAPTER 1

#### PRELIMINARY DEFINITIONS, EQUILIBRIUM DIAGRAM

**Introduction.** Steel combines strength, workability, and cheapness to a degree unparalleled in any other material of construction. It is the mainstay of engineering, indispensable alike in peace and war.

The engineer would be badly cramped in the utilization of steel were his use of it confined to those purposes sufficiently well served by steel as it comes from the rolls, the forge, or the mold. Improvement in and gradation of the properties of steel, conferred by controlled heating and cooling, fit heat-treated steel for myriads of uses which raw steel will not satisfactorily serve.

Finer and finer gradations of properties are being and will continue to be demanded by the user of steel. Still closer control of steel-making processes and of heat-treating methods must be developed. To this end accumulated practical experience, improved equipment, and increased understanding and application of the fundamental principles involved must all be utilized. Men, machinery, and knowledge make a stool which needs all three of its legs to stand on.

Vast strides have been made in recent years in the way of equipment. Processes that were previously workable only in the laboratory by application of the greatest skill have become the commonplaces of production through the development of special equipment and of control devices, far more than human in their action. Continuous and automatic processes are in vogue that perform operations with a speed and certainty that far surpass the skill of the artisan and maintain a uniformity of quality unthought of when only manual dexterity was involved. But the planning of such processes involves exactly the same metallurgical principles as were utilized in nonmechanized processes. The human element is still involved. Precise knowledge is even more emphatically demanded than it was when the properties of the product were not so rigorously specified.

## EVALUATION OF STEELS

Certain conventional and many special tests are applied to evaluate heat-treated products, and to appraise the suitability of a type of steel, or of a given lot of steel, for treatment. Whether or not the heat treater himself makes these tests, he needs to understand how they are made and how to interpret them. The limitations of the tests need especial attention, for good material may be excluded when a test is applied, which offhand *seems* to indicate quality, but which really does not, because the results are not actually pertinent to the case in hand.

The simple conventional tests of properties never evaluate all the attributes that are really required, but they are applied on the supposition that, if, in the attributes determined, the samples agree with the attributes of materials previously and satisfactorily used in the same or similar service, the attributes not examined will also prove satisfactory. This supposition is not always verified. However, wide variation, in the conventional tests, from the expected values, at once raises a suspicion that the important but unmeasured attributes that really count would prove unsatisfactory. This suspicion is not always verified, either. Nevertheless, the conventional tests serve as a screen to separate out the cases where the engineer feels justified, by virtue of likeness shown, in using the material without exhaustive service experience, from the other cases where he will demand simulated service tests before approving the material.

The conventional tests provide a way in which some requirements can be given in exact terms rather than in vague phrases. The fewer the attributes that are really needed, but are unmeasured (or unmeasurable), the simpler it is to specify and to produce material that will truly meet the requirements of service. Hence much more testing of steel for heat treatment, as well as of the finished product, is being done than of yore. This is a step in advance, as long as only the truly applicable tests are made.

With this increase in testing and inspection, it becomes necessary to make every test count. Supposedly "duplicate" tests do not always check exactly, and the degree of variation outside of which the "duplicates" must fall before they really indicate differences in quality is a matter of real importance in making the language and terminology of tests and specifications precise and understandable.

As is true of every specialized field, heat treating has a language and terminology of its own, adapted both for brevity and for exactness.

Before taking up steel and its heat treatment, it is advisable to define and explain the terms and concepts used in evaluation of steel and in all discussions of heat treatment. Many definitions can be left until the discussion requires them, but some definitions and some discussion of the aims and methods of mechanical testing should precede detailed consideration of heat treatment.

### HEAT TREATMENT

"Heat treatment of steel," in its broadest sense, refers to any process involving heating and cooling of the solid metal by which the properties of the steel are altered without any intentional alteration of its chemical composition.

Carburization, which alters the chemical composition, is often loosely included in the definition, but the true heat treatment of carburized steel is performed after the chemical change has occurred. Decarburization, or scaling of the surface, may accompany some heat-treatment operations, and great precautions may be required to avoid these, usually undesirable features.

In a more limited sense, the term heat treatment applies especially to the quenching and tempering (hardening and toughening) processes, so that the term "heat-treated steel," without further qualification, is applied to steel that has been cooled with a certain degree of rapidity to produce changes in its structure.

In a still more limited but most widely used sense, it applies to cooling at such a rate that the hard brittle material called "*martensite*" is formed, followed by a tempering reheat to toughen and generally to soften it. The initial production of martensite is a primary aim of heat treatment of this type. However, the single operations of cooling at a specified rate, as in "normalizing," or of heating, as in the softening of a cold-worked steel by annealing, come under the broad definition.

The object of heat treatment is to produce mechanical or physical properties that make the steel better adapted for industrial use.

**Steel, Alloy Steel.** Steel is defined as an alloy of iron and carbon, usefully malleable as cast. More specifically it is an alloy of iron and iron carbide, which is a compound of iron and carbon.

The phrase "malleable as cast" serves to exclude gray cast iron, white cast iron, and malleableized cast iron. For practical purposes one might take the lower carbon limit of the low-carbon steel, "1010," that is, 0.08%, as the dividing line between iron and steel.

The lower limit of the carbon content might alternatively be fixed as that value below which iron carbide cannot be thrown out of solid solution at any temperature. This is lower than 0.01%, and so practically all commercial irons are to some slight degree amenable to heat treatment and could be considered as steels. Common parlance tends to call the very low-carbon alloys, that are only slightly or with difficulty affected by heat treatment, "irons" rather than "steels."

The practical upper limit is not clearly defined. Save for tool steels, there are few steels that contain more than 1% carbon, and in these 1.25% is about the upper limit. The "graphitic" silicon steels may have as much as 1.50%. From the point of view of theory, an upper carbon limit of 1.7% (or 2.0%, see legend of Fig. 3), corresponding to that point in the iron-carbon equilibrium diagram, Fig. 3, beyond which iron carbide cannot be wholly held in solid solution at any temperature, is often linked with the definition of steel. Although common terminology disregards this limit in some alloy steels that have a different type of carbide, for example, high-carbon high-chromium steels, the limit does serve as a first approximation.

**Commercial Steels.** Actually, a truly pure alloy of iron and carbon alone would behave quite differently from a commercial carbon steel of equal carbon content because of the intentional addition to true steel of manganese, silicon, or some other "deoxidizing" element, added to counteract the tendency for iron oxide to be present along with iron and iron carbide. These added elements are considered as normal constituents of carbon steel until their amounts are raised to such proportions that deoxidation is no longer the main purpose and some specific effect is obtained by the addition, when an "alloy steel" results. Traces, at least, of sulphur and phosphorus and generally of other elements, not intentionally added but arising from the iron ore and the scrap used as raw materials, will also be present. A very little aluminum is often present, for grain-size control, in modern steels for heat treatment. Small amounts of other "deoxidizers" may be used in place of aluminum for this purpose. Such steels are still classed as "commercial carbon steels."

The definitions of carbon and alloy steel adopted by the American Iron and Steel Institute are drawn up on the basis of commercial pricing rather than on that of differentiating between use of elements for straight deoxidation and their use for specific strengthening and alloying effects. The AISI definition of carbon steel is one in which no definite amount of other common alloying elements is required (though small residual amounts coming from scrap are not barred), and in which the maxima for manganese, silicon, and copper are, respectively, 1.65%, 0.60%, and 0.60%.

From our present point of view, a steel of 1.65% manganese is definitely an alloy steel, responding very differently to heat treatment from one in which only sufficient manganese is present for deoxidation; and, from the point of view of atmospheric corrosion, considerably less than 0.60% copper definitely makes an alloy steel. Reasonable upper limits for manganese and silicon in what one might regard as unalloyed steels might be taken as 0.50% and 0.35%, respectively. In steel castings, however, the silicon content runs higher, and 0.50% would be a more natural dividing line. Several of the standard "carbon steels," indeed, those with which the heat treater is most concerned, carry around 0.60–0.90% manganese, a few even higher, and so consideration of carbon steels cannot be fully untangled from the alloying effect of manganese, since this element is so universally used.

#### ELEMENTS OCCURRING IN STEEL

At least 25 of the chemical elements are usefully employed in steels of different kinds. In metallurgical as well as chemical literature these elements have to be mentioned constantly. Their names are abbreviated to the chemical symbols. Since the reports of the chemical analyses are universally given without the full name of the element being spelled out, the abbreviations must be learned.

These abbreviations are hereafter used in the text in order to save space. Following are the names of and symbols for the steel-making elements:

Iron	Fe	(Latin, ferrum). The base of all steels.
Carbon	C	A nonmetal. The element which most influences steels in respect to heat treatment.
Silicon	Si	A small amount is present in all ordinary killed * steels.
Manganese	Mn	A small amount is essential in all ordinary steels.

\* See p. 67.



6      *PRELIMINARY DEFINITIONS, EQUILIBRIUM DIAGRAM*

Phosphorus	P	A nonmetal, but alloys with iron.	
Sulphur	S	A nonmetal, present as sulphides.	
Oxygen	O	A gas, present as oxides.	} Both may also be present as the elements, but only in a tiny amount.
Nitrogen	N	A gas, present as nitrides.	
Hydrogen	H	A gas, present, probably as the element, in most steels, but only in a very tiny amount.	

The elements listed are present in so-called "plain carbon" steels. The amount of Si, Mn, P, or S may be increased beyond the normal in order to confer special properties, when they may be classed as alloying elements.

The other important metallic alloying elements are:

Chromium	Cr
Copper	Cu (Latin, cuprum)
Nickel	Ni
Molybdenum	Mo
Tungsten	W (Latin, wolfram)
Vanadium	V

Other metals used in special alloy steels or added for particular purposes are:

Aluminum	Al
Bismuth	Bi
Boron	B
Cobalt	Co
Columbium	Cb
Lead	Pb (Latin, plumbum)
Silver	Ag (Latin, argentum)
Tantalum	Ta
Titanium	Ti
Zirconium	Zr

Nonmetals occasionally used for purposes similar to those when S is utilized are:

Selenium	Se
Tellurium	Te

A few more elements may accidentally occur in or be intentionally added to steel, but in such small amount or so rarely as to need little comment from the point of view of the heat treater. Among these are antimony, Sb (Latin, stibium); arsenic, As; tin, Sn (Latin stannum); Sn comes in from incompletely detinned scrap; As and Sb from certain ores. All three may enter from badly sorted automotive-engine scrap. None of these is easily eliminated.

Zinc (Zn) is often charged to the steel furnace as galvanized scrap but boils out almost completely and so does not figure in the finished steel.

Calcium (Ca) is a constituent of some complex addition agents used for control of grain size and hardenability, but it also boils out almost completely. When present, it is in the nonmetallic inclusions.

Nearly every other metal in the periodic system has been added to steel. Uranium (U) might serve somewhat as do Mo and W, and was tried to some extent in high-speed tool steel long ago, a practice since abandoned. Beryllium (Be), copper (Cu), and titanium (Ti) are capable of precipitation-hardening effects in steel; cerium (Ce) has some power to combine with S.

In connection with composition, 0.01% of an element (except S and P) is called a *point*. Steel A 1012, with 0.10 to 0.15% C, has a six-point range. In the case of S and P, 0.001% is called a point.

## RESIDUAL METALS

From a practical point of view, important metallic elements are introduced into "C steel" through scrap. The "residual" metals which come from scrap and are retained in the finished steel are Cu, Ni, Cr, Mo, Sn, As, and Sb. A survey<sup>1</sup> of C steel produced in the United States and Canada in the last half of 1944 yielded the data given in Table 1.

TABLE 1

	<i>Grand Average,</i> %	<i>High Average for Individual Plants,</i> %
Cu	0.10	0.22
Ni	0.05	0.17
Cr	0.09	0.09
Mo	0.015	0.03
Sn	0.015	0.04

To reach these averages some individual heats were, of course, much higher. As a rule steel carries also an average of 0.01% As and 0.01% Sb.

Small traces of V, Al, Ti, and Zr may remain from previous treatment of C steel in the scrap with alloys containing them, but these are more likely to be present, after the scrap is melted, as sulphides, oxides, nitrides, or silicates, that is, in inclusions, rather than as metal. (Intentional additions of one or more of these four, of the general order of 0.05%, are often made to C steel.) Alloy scrap, of course, contains material amounts of the alloying metals employed.

<sup>1</sup> Bibliographic references are at the end of each chapter.

Much higher levels of residuals may be met in mixed C and alloy scrap, and, when a high level occurs, it is often profitable to add more alloy to bring the heat to an acceptable "N.E." (National Emergency) alloy composition, rather than to let high residuals appear in what is to be marketed as "C steel." Rapid spectroscopic analysis of the melt for residual elements makes this feasible.

Nelson<sup>2</sup> estimated in December 1945 that each 0.10% of the following alloying elements in scrap was then worth, per net ton of ingots:

Mn	\$0.16
Cr	0.24
Ni	0.60
Mo	1.60
V	5.40

The level of residuals is of some concern when very soft steels are being made to be used for deep drawing and like processes involving severe deformation, for the elements all have more or less a strengthening effect, not wanted in this case. When several residuals are at a high level, their effect may even be felt in the higher-C heat-treatable steels and may account for differences in hardenability, particularly in low-alloy shallow-hardening steels. Dependence on routine chemical analysis that does not determine the residuals, fails to take account of such variations. However, when proper hardenability determinations are made, the effect of residuals that would affect heat treatment is thereby summarized.

#### STANDARD COMPOSITIONS, NUMBERING SYSTEMS

Inasmuch as a discussion of principles must often refer to other than plain C steels, and both these and the most-used commercial alloy steels have been given numbers which serve as shorthand in referring to composition (a practice first instituted by the Society of Automotive Engineers); the AISI (American Iron and Steel Institute) and N.E. (National Emergency) numbers are given on pages xxvi-xxvii in the front of this book. In this system the prefixed letter indicates the melting process, the first two digits signify the class of steel, and the final digits the C content.

**"H" Steels.** Purchase of steels for heat treatment is increasingly being made by specification of hardenability upon quenching, this property usually being indicated by the Jominy end-quench test hereinafter discussed. Certain hardenability limits are set up, and

a steel guaranteed to be within those limits is obtained by ordering, say, 8630 H; whereas an order for 8630 will be filled with a steel meeting the chemical specifications, but not necessarily the hardenability limits.

In 1945, the following had been designated as "H" steels and the hardenability bands for each published.<sup>8</sup>

- 2512, 2515, 2517
- 3310, 3316
- 4130, 4132, 4135, 4137, 4140, 4142, 4145, 4150
- 4317, 4320, 4340
- 4620, 4640
- 4815, 4820

In the 8600, 8700 and 9400 series, the 20, 22, 25, 27, 30, 32, 35, 37, 40, 42, 45, 47, and 50 members, for example, 8620, 8450, were all included. By 1947 some of these were dropped from the AISI list and hardenability bands given for others. Ultimately, hardenability bands will be published by AISI for all commercially procurable steels designed for quenching and tempering.

**"Treated" Steels.** The steels now designated A 1330 to A 1350 were earlier designated by T 1330 to T 1350. The "T" is now employed to denote a "treated" steel, that is, one which has had an addition of ferrobore or of a complex "deoxidizing" alloy, containing B, as well as some combination of Si, Al, V, Ti, Zr, Ca, to add quench hardenability. The T is inserted between the first and second pairs of digits. A "treated" 1330 or 4640 is designated 13T30 or 46T40. "A46T40H" thus refers to 4640 made in the basic open hearth (A), B treated (T), and supplied to the hardenability specification (H).

The 71000 and 72000 series are W-Cr, tool and die steels, the 30000 and 51000 series are corrosion- and heat-resisting alloys to which it will not be necessary to refer in this volume.

Some other numbering systems are cluttered up by a variety of numbers for the same steel. The AISI designations conform to the easily remembered SAE system, but the AMS (Aeronautical Materials Specifications) numbers do not always do so, for example:

AMS	{	5025 is 1137
		6254B is 3312
		6290B is 4615
		6352 is 4130
		6415 is 4330
		6440 is 52100

Some of these can be spotted as AMS numbers because they cover 5000, 6200, 6300, and 6400, which are not in the AISI-SAE series, but the 5100 series is liable to confusion, for example:

5120 AMS is 0.70% C sheet spring stock of 1070 steel.

5120 AISI is 0.20 C, 0.80% Cr.

Pending better unification, when AMS designations are used, it should be carefully stated that they are AMS rather than AISI numbers. Cross indices to the various designations, including the wide variety used by various Government departments, are available.<sup>4, 5</sup>

The N.E. (National Emergency) steels were promulgated in the period of war shortages as a conservation measure in order to utilize the Ni, Cr, and Mo contents of mixed alloy scrap. When alloys needed to be added to achieve a composition of a higher level of hardenability, Mo, available from domestic sources, expensive, but very potent, was often utilized. Mn and Cr, of which stocks were available and domestic production, though insufficient in itself, had been stepped up, were also used. Special effort was made to avoid addition of Ni, of which the supply was short. Addition of V was likewise avoided, since the V supply was needed for high-speed tool steel.

The British had similar difficulties because of scarcity of alloying elements and met them in much the same way as we did on this side of the water. Purchase by hardenability instead of by chemical composition was favored. The number of compositions regularly produced and stocked was reduced, and the simplified series was given "E.N." numbers. A few low-alloy compositions, analogous to the American N.E. compositions, were added. Among these, E.N. 100 corresponded fairly closely to our N.E. 9400 series.

The philosophy back of the E.N. steels was set forth by Hatfield,<sup>6</sup> and their compositions, properties, and usable sections were discussed in detail by Monypenny.<sup>7</sup>

Considerable technical literature appeared concerning the properties of and the substitution of some of the early N.E. compositions for SAE compositions, when these were first put into production, before it was concluded that these compositions were superfluous, or did not lend themselves so well to conservation as others. Emphasis was at first placed on the 9200 high-Si series because of cheapness and availability of Si, but mill difficulties in handling such steels soon relegated them to a rather minor position.

From the point of view of principles, the data are as valuable as any others.

PRELIMINARY DEFINITIONS RELATING TO THE  
EQUILIBRIUM DIAGRAM

Just as the chemical symbols serve as a convenient shorthand, so certain words and phrases are used in metallurgy and heat treatment to convey definite ideas which otherwise would require long-winded explanations to be expressed clearly. The definitions and terminology are repeated and enlarged on, or more fully explained by the context, at suitable places in the discussion that follows.

However, it is often necessary to use a word before it has thus been defined in detail; hence there are here collected some preliminary explanations.

*Ferrite* is practically pure iron. It is crystalline (Fig. 1) with what crystallographers call a *body-centered cubic structure* (Fig. 2). This is also called *alpha iron*. It exists at room temperature and up to 1670° \* in the absence of C, but, as C increases, its upper limit of existence is lowered to about 1335°, at about 0.80 to 0.83% (and higher) C. It has very little power to dissolve C. It is *magnetic*. Above these temperatures it changes to nonmagnetic *austenite*, which has another crystalline-structure *face-centered cubic* (Fig. 2). This is also called *gamma iron*. Austenite has a relatively high solvent power for C. All we need here be concerned with is the fact that alpha and gamma are different.

Both types of crystals are in the cubic system. This information is gained by X-ray crystal-structure methods of study. The terms body-centered and face-centered are common in modern metallurgical literature. Since in ordinary steels one brings the metal into the austenitic condition by heating in order to cool it at a controlled rate, one can remember that austenite is face-centered since it is made ready for cooling. Another way is to recall that the face is above the body, and austenite is above ferrite in the Fe-C diagram (Fig. 3).

The shift from one crystal form to another is called *transformation*. The temperature at which the shift occurs is called a *critical temperature* or *critical point*. The "points" are usually spread over a range of temperature, and are, therefore, also called "*critical ranges*." The critical temperatures at which Fe changes are altered by the presence of C.

\* All temperatures herein are given in degrees Fahrenheit. A conversion table to centigrade appears on p. xx in the front of the book.



FIG. 1. Slowly cooled iron, deeply etched to show crystalline structure. Etched in 9%  $\text{HNO}_3$ .  $\times 900$ . (Dept. of metallography, Univ. of Minn.)

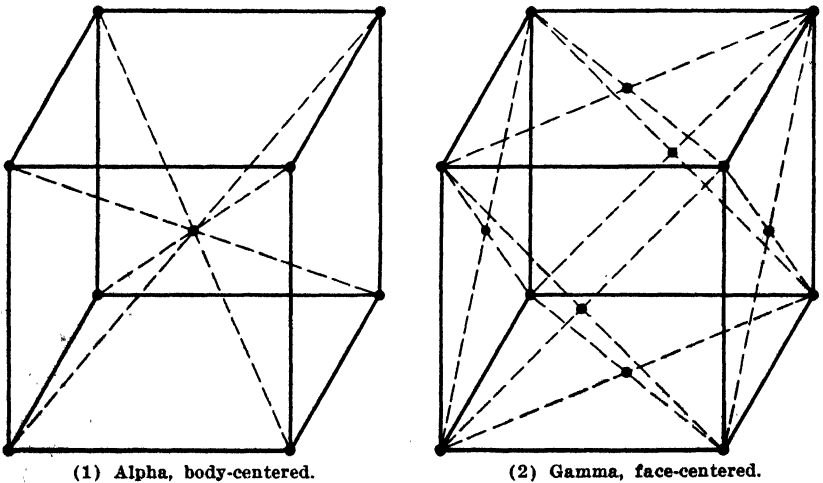


FIG. 2. Arrangement of atoms in the allotropic forms of iron.

Each different crystal structure is called a *phase*. Austenite is a single phase. It can contain C but it still has the same crystal struc-

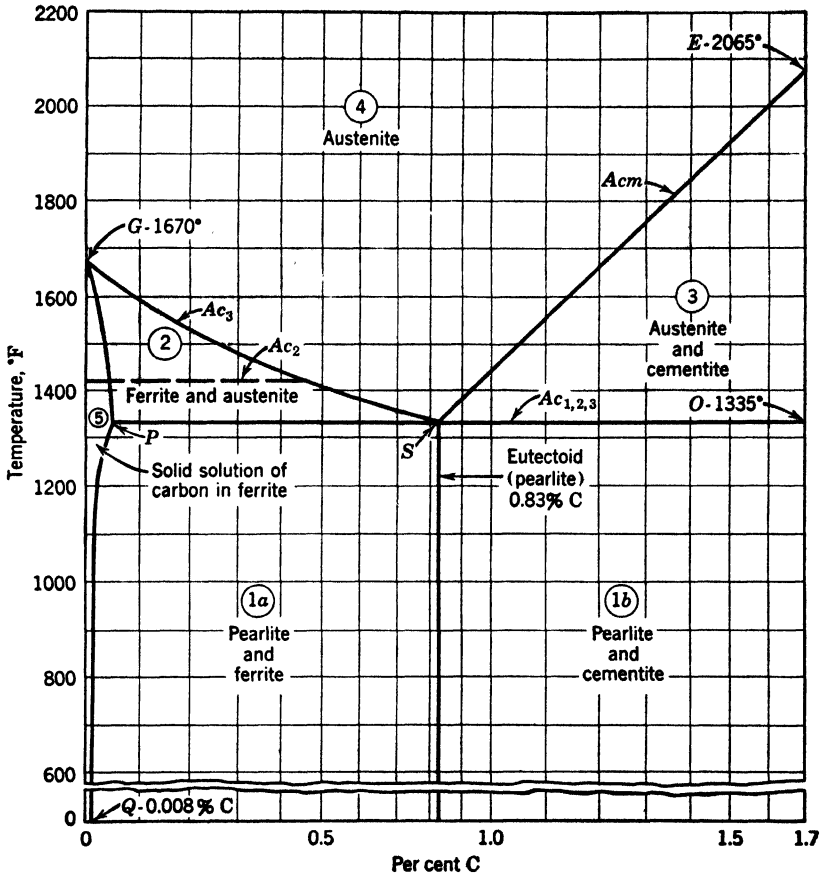


FIG. 3. Iron-carbon equilibrium diagram.  $Ac_1$  occurs at the line  $PS$ . The lines  $GS$  and  $SE$  bounding the austenite field are of primary importance in the heating phase of heat treatment, and line  $PS$  in the cooling phase. The eutectoid composition, here shown as 0.83% C, and the point  $E$ , denoting the maximum solubility of carbon in austenite, here shown as 1.7%, are given by the committee revising the diagram for the 1947 *Metals Handbook*, as 0.80% C and 2.0% C, respectively; <sup>8</sup> see Fig. 3b.

ture, regardless of its C content. The spacing of the atoms within the "crystal lattice" increases as the C content increases, but the Fe atoms still have the same arrangement.

A map of the temperatures at which these phase changes occur on very slow heating or cooling in relation to C content is called the



Fe-C equilibrium diagram (Fig. 3). Figure 3a shows the temperature of various heat-treatment operations in relation to the equilibrium diagram. Figure 3b expresses the latest beliefs as to the positions of the points designated S, E, and P.

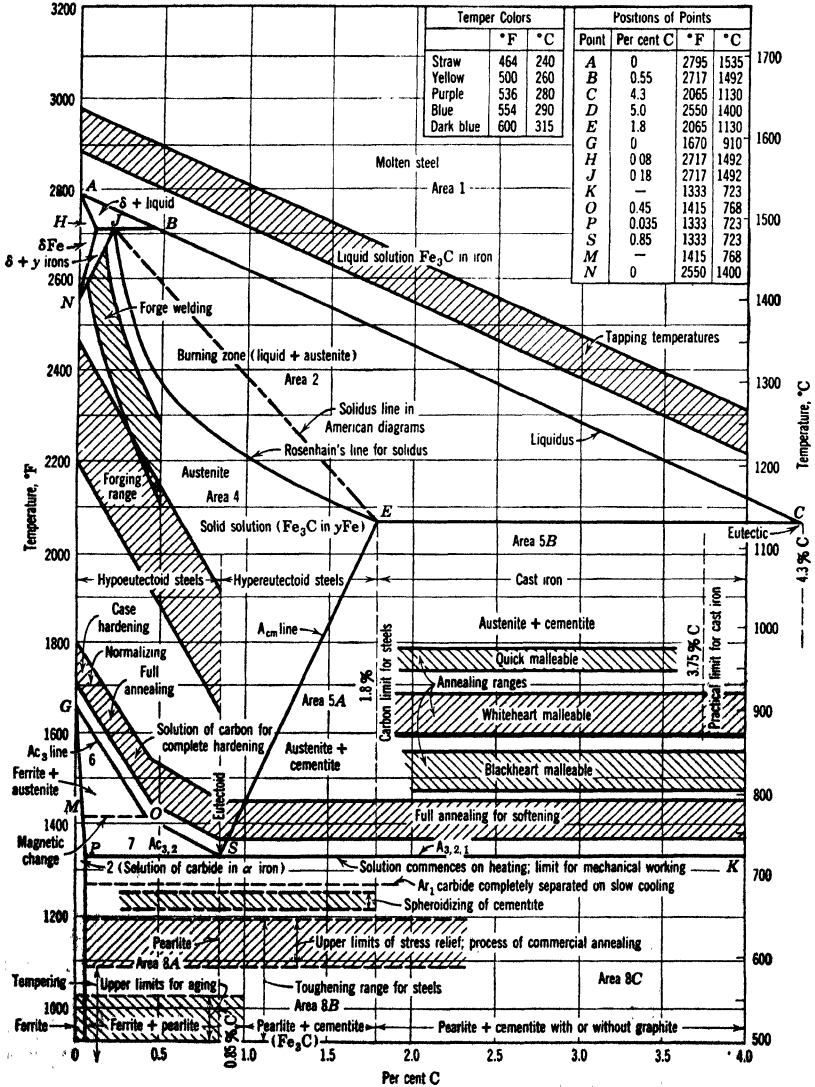


FIG. 3a. Iron-carbon equilibrium diagram. Modified for students and shopmen by R. Whitfield. (From *Metal Progress*, December 1945, p. 1312B)

Although the equilibrium diagram is insufficient to explain what happens in heat treatment by quenching and tempering, it is a necessary preliminary to such an explanation.

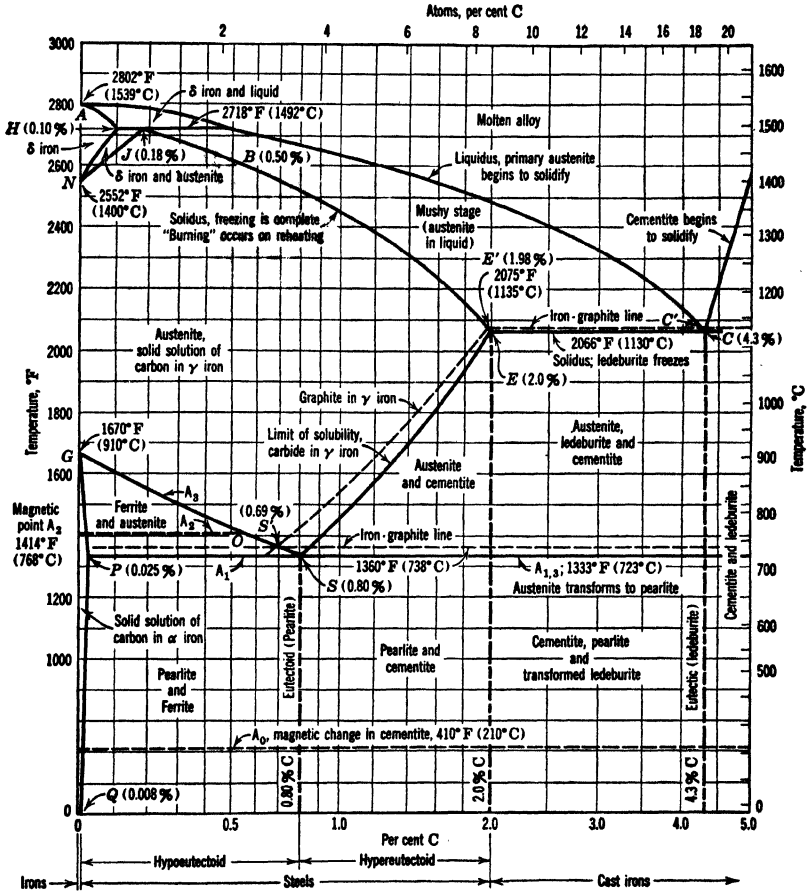


FIG. 3b. Iron-carbon equilibrium diagram, revised 1946. Principal points and lines, especially *GSEJ*, adjusted from positions in former edition of *Metal Progress* data sheet, to agree with diagram prepared for the subcommittee on phase diagrams, Kent R. Van Horn, chairman; John S. Marsh; and F. N. Rhines; for inclusion in the 1947 *Metals Handbook*. Point Q, according to J. H. Whiteley, *Journal of the Iron and Steel Institute*, 1936. (From *Metal Progress*, November 1946, p. 970)

The nonferrous metallurgist obtains much of his basic information about alloys from equilibrium diagrams. The steel metallurgist does likewise, but in his case that information is far from sufficient. Equi-

librium diagrams represent behavior under conditions of extremely slow heating and cooling. For those conditions, they map out the fields of composition and temperature in which various constituents are stable, and show the fences between the fields, the boundaries at which the constituents change or transform.

No provision is made by such diagrams for the case where, by rapid cooling, a constituent climbs the fence and gets over into a field where it does not belong and where it cannot stay permanently.

Steel not only can climb the fence; it can jump over it. The transformations that occur in steel when the fence is slowly climbed are suspended when it jumps over, but tend to occur with more readiness the further the foreign territory has been invaded.

Moreover, steel can undergo a transformation, that to hard *martensite*, for which there is no field whatever on the equilibrium diagram. Indeed, the only type of heat treatment which the equilibrium diagram describes with any degree of accuracy is the operation of full annealing, complete softening. None of the types of treatment by which hardening and strengthening of steel are accomplished through use of rapid rates of cooling are described by it.

However, as a mapping of behavior on slow heating and cooling, the equilibrium diagram for Fe and C, the primary elements in steel, is the starting point for any study of heat treatment. It shows one constituent, austenite, whose fence-jumping proclivities make regulation heat treatment possible; and with which discussion of heat treatment is continually concerned.

Although the equilibrium diagram is for Fe and C, in the cold steel that has reacted as the equilibrium diagram says it has, there is no C present, save in the rare case of graphitization. The C is in the combined form, in a definite chemical compound containing three atoms of Fe to one of C; it is  $\text{Fe}_3\text{C}$ , iron carbide (cementite) with 93.4% Fe, 6.6% C. Cementite is named from the old "cementation" process in which wrought iron (relatively pure ferrite) was heated with C to produce steel.

There are *two* constituents in cold (annealed, normalized, or hot-rolled) steel, cementite and alpha iron. When the steel is heated high enough, there is only one constituent, austenite.

Pure Fe changes from ferrite to austenite on heating past  $1670^\circ$ ; on cooling below  $1670^\circ$ , the austenite changes back to ferrite.

The diagram, Fig. 3, shows that, at high temperatures, in spite of the previous presence in the cold steel of two constituents, Fe and its

carbide, there is only one constituent. To get one out of two, one of the two must swallow the other.

Just as water dissolves salt, making brine, so does gamma iron dissolve iron carbide, the only difference being that the gamma iron is solid instead of liquid, so that the product is a "solid solution."

Just as brine may contain different amounts of salt in solution, so austenite can contain little or much C. Its solvent power for  $\text{Fe}_3\text{C}$  is, however, limited, just as that of water for salt; and, again, as in the case of brine, increases with temperature, as may be seen at the upper right of Fig. 3. Hence, in the field to the right of the line  $SE$  there are two constituents, austenite and cementite.

The point  $S$  in Fig. 3 is one of special interest. It lies on the horizontal line corresponding to  $1330^\circ$  or  $1335^\circ$ , and on the vertical line corresponding to about 0.80 or 0.83% C (see legend of Fig. 3). A C-rich austenite, say one of 1.5% C at  $2000^\circ$  cooling from that temperature, remains the same until it gets just below  $1925^\circ$ , when, the solubility limit for carbide at that temperature having been reached, cementite separates out, just as salt separates from cooling brine. The austenite, thus depleted of enough of its original carbon to correspond with its new temperature, continues to throw out more and more as the temperature approaches  $1335^\circ$ . At that temperature iron carbide corresponding to about 0.67% C, out of the original 1.5%, has been thrown out as cementite, and approximately 0.83% remains. The austenite has slid down the line  $ES$ . At  $S$  the Fe of the austenite transforms from the gamma to the alpha form, in which latter case the solubility of C is low (as shown by the location of the point  $P$ , and so there can no longer be a solid solution containing more than about 0.04% C.\* So the remainder, 0.79% C, comes out with a rush, forming cementite. These two things happen nearly simultaneously, but one slightly precedes the other, since the separated alpha ferrite and the separated cementite, under slow-cooling conditions, come out in layers, plates of cementite being interleaved between layers of ferrite. This layer or "lamellar" structure has a "pearly luster" and is called *pearlite* (Fig. 4).

Before Al additions became common in the finishing of a heat of steel, this lamellar type of pearlite was the type ordinarily used. In Al-treated steels the pearlite may be wholly or partly of a nonlamellar structure, with the carbide distributed as tiny islands rather than in

\* More recently<sup>8</sup> this figure has been taken as 0.025%, as in Fig. 3b.

strata. Grange<sup>9</sup> has given the name *semipearlite* to such a structure. It is not necessary to differentiate between two types at this point in the discussion.



(a) Optical microscope.  $\times 1000$ . (Elsa, Battelle)



(b) Electron microscope.  $\times 12,000$ .

FIG. 4. Pearlite and cementite plates in layers of ferrite. (a) At 1000 diameters, ordinary microscope. (b) At 12,000 diameters, electron microscope, using Formvar replica, shadow-cast with gold. (Schwartz, Battelle)

Our 1.5% C steel now consists of the separated cementite that came out as the austenite followed the line *SE* down, in a matrix of the pearlite formed at 1335°, that is, at the point *S*.

If we follow down the cooling of austenite of a lower carbon content, say 0.30% carbon, starting at 1600°, nothing happens until it cools to about 1440°. This time ferrite separates, instead of car-

bide, and the austenite becomes poorer in Fe and richer in C. It slides down the line  $GS$ , continually losing ferrite and gaining in concentration of C, until the C content of the remaining austenite has built up to 0.83% and the temperature has fallen to 1335°; that is, the point  $S$  has been reached. On further cooling, the 0.83% C austenite separates into pearlite, just as it did when that carbon content and temperature were approached from the right. The 0.30% C steel now consists of the alpha ferrite thrown out along the line  $GS$ , and the pearlite aggregate of ferrite and cementite. Reverse changes occur on heating (Fig. 5).

**“Beta” Iron.** The diagram, Fig. 3, shows a dashed horizontal line crossing the ferrite fields at about 1420°, at which temperature ferrite becomes magnetic on cooling, or loses its magnetism on heating.\* But the crystal structure is the same above or below 1420°. There is no transformation. In earlier days, magnetic ferrite, below 1420°, was called alpha iron; nonmagnetic ferrite, above 1420°, beta iron. This explains why the iron in austenite, produced at a still higher temperature, was termed gamma iron instead of beta. The term beta iron is now used merely to imply hot nonmagnetic alpha, not a different ferrite.

If the diagram of Fig. 3 were extended upward to include melting and freezing, on the pure Fe side, there would appear a little field adjacent to the corner of the austenite field in which ferrite again appears; this is termed delta ferrite. It is shown in Figs. 3a and 3b. However, delta is no different from alpha. Neither beta nor delta ferrites have anything to do with ordinary heat treatment. For our present purposes we can forget all about them.

**The Eutectoid.** The point  $S$  represents the lowest temperature at which (under very slow cooling) gamma iron can exist, also the one composition, about 0.80 or 0.83% C, at which austenite transforms

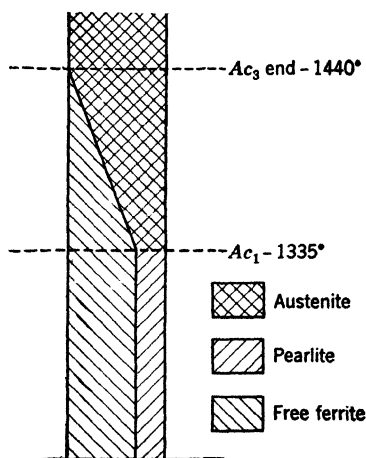


FIG. 5. Change of pearlite and free ferrite into austenite during heating or slow cooling of 0.30% C steel.

\* This temperature is not exact, but is that at which the loss of magnetism is most rapid. Some magnetism is lost a bit below 1420°.

wholly to pearlite without throwing out either alpha ferrite or cementite. This lowest-temperature constant-composition point, in a binary equilibrium diagram showing liquid to solid transformations, is called the eutectic, from Greek words meaning "easy melting."

On account of this analogy, the point *S* is called the "eutectoid," the term used for the transformation of a solid to two solids, as contrasted with liquid to solid.

Hence steel of 0.80 or 0.83% C, whichever figure one prefers to accept, is "eutectoid steel." Steel with less C is *hypoeutectoid* (hypo means under) and that with more is *hypereutectoid* (hyper means over). The ferrite thrown out, as austenite cools and slides down the line *GS*, is "*proeutectoid ferrite*"; the cementite thrown out as it cools and slides down *ES* is "*proeutectoid cementite*" (pro means ahead of or prior to).

In *pure* Fe-C alloys the eutectoid not only would be fixed as to C content, but also would occur sharply at 1335°. In commercial steels, containing Mn, Si, etc., the C content of the eutectoid is altered somewhat (usually being less than 0.83%), and the transformation occurs over a small *range* of temperature instead of exactly at 1335°. Elements that readily form carbides, such as Cr and Mo, sharply raise the eutectoid temperature; whereas Mn and Ni lower it appreciably.

**Solubility of Carbon in Ferrite.** One other field calls for comment, the field *GPO*, for alpha ferrite containing a little C, about 0.04% or less at 1335°, down to around 0.008% at room temperature. The ferrite thrown out from austenite at 1335°, or above, should, under slow-cooling conditions, contain this 0.04% C at 1335°, and on cooling below that should reject 0.032% C, sliding down the line *PO*. The line *PO* is curved in a way that is very familiar to non-ferrous metallurgists, since such a curvature is found in the equilibrium diagrams for many "precipitation-hardening" systems. As is discussed in Chapter 10, ferrite is amenable to precipitation hardening by separation of its dissolved C, tiny though the amount of C is. This phenomenon is of minor effect in most heat-treatable steels of much high C content, the effects being there overshadowed by those resulting from other transformations. These other transformations can be discussed without attention to what happens, owing to the rejection of 0.032% C.

**Ac and Ar.** There are other notations on Fig. 3 which require comment. If one puts a thermocouple in a piece of steel and with sufficiently sensitive equipment records a time-temperature slow-heating curve, there will be a break in the curve at 1335°. This distortion

of the curve remains until the line *GS* or *SE* has been passed. There is an "arrest" in the curve. Heat is required to provide energy for the transformation of alpha to gamma iron and for the solution of C in Fe. The French for heating is "chauffage." So the heating arrest is designated "*Ac*" (*A* for arrest; *c* for heating). This is the "critical point" on heating. Although the heat requirement at the magnetic change point at 1420° is tiny, three stages can be noted in hypoeutectoid steels, first *Ac*<sub>1</sub>, transformation of pearlite to austenite of 0.83% C; *Ac*<sub>2</sub>, magnetic change in the ferrite at 1420°; and *Ac*<sub>3</sub>, representing the transformation of the last remaining ferrite to austenite. In a eutectoid steel, once the pearlite has transformed to austenite of 0.83% C at 1335°, there is no more ferrite, and yet the transformation is often termed *Ac*<sub>1-2-3</sub> to indicate that all stages of the transformation are merged into one.

In hypereutectoid steels, the detection of the change in direction of the time-temperature heating curve as the line *SE* is crossed and the last trace of cementite is dissolved is difficult, and so the exact temperature of the *Ac<sub>m</sub>* (cementite arrest) point is not easy to determine, its location being more often taken by reference to the diagram, or by heating, quenching, and microscopic examination.

On cooling, heat is given out as the lines *GS* and *PS* or *ES* and *PS* are passed. The temperature may drop slightly below the transformation temperature without heat evolution. This is supercooling. When the temperature drops a little further, the supercooled material transforms rapidly, with the development of the delayed heat evolution, the temperature then rises, quite notably at the 1335° point. This is termed "recalescence," and the transformation temperature for this "critical point" on cooling is sometimes called the "recalescent point." The arrests on cooling are designated "*Ar*," since the French word for cooling is "refroidissement." (To keep the distinction between *Ac* and *Ar* in mind, the English words caloric for heating and refrigeration for cooling may be recalled.)

**Homogenization.** As was stated before, the equilibrium diagram only tries to map the fields that exist when heating and cooling are extremely slow. There is a time factor attending the attainment of equilibrium on heating. It is possible (especially with electrical resistance or induction heating) to flash the temperature up so fast as to "climb the fence." At any rate, we must remember that in heating, say, a 0.30% C steel, at 1335°, the pearlite alone changes to austenite, the free ferrite (that not in layer arrangement interleaved with cementite) has still to be absorbed; on very slow heating the



line *SG* would be climbed, the austenite that came from pearlite steadily picking up more and more ferrite till at 1440° the austenite that came from pearlite has picked up all the ferrite and reduced its own C content from 0.83% to 0.30%, as indicated in Fig. 5. But in very rapid heating the solution and diffusion cannot keep up, and the austenite from pearlite with 0.83% C can get above 1440° without having its C content much reduced. The structure will contain patches of nearly 0.83% C austenite and patches of nearly carbonless austenite from ferrite. Higher temperature or long time just above 1440° would be needed before the austenite became homogenized and everywhere had a uniform 0.30% C. Such *homogenization*, or its lack, has a vital influence on heat treatment, as we shall frequently see in later discussion.

The attainment of full solution of cementite at *Acm*, on the line *SE*, in hypereutectoid steels, is even more difficult. Indeed, full solution is not always attempted. Heating only between *SE* and *SO* is often done intentionally. In hypoeutectoid steels, however, few useful results are obtained by heating only between *GS* and *PS*. The austenite of hypoeutectoid steels is almost invariably heated above the line *GS*, and time enough given for solution of ferrite.

While rushing the *heating* thus has some effect, equilibrium is established without very great difficulty. It would indeed be hard to retain much ferrite on heating above the lines *GSE*, though complete homogenization of the austenite goes on more leisurely.

But, when *cooling* is rushed, conditions are quite different. The steel gets stiffer as it cools, and reactions tend to go on more slowly. It is therefore possible to retain austenite below 1335°, and thus lower the temperature and delay the time of its transformation. Indeed the retention of austenite and its decomposition at a controlled temperature *lower* than 1335° is the heat treatment of steel in a nutshell.

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See pp. xxxi-xxxii for abbreviations of the names of technical societies and for their addresses.

## CHAPTER 2

### NONEQUILIBRIUM DEFINITIONS AND CONSIDERATIONS. COOLING CURVES. SPLIT TRANSFORMATION

The temperature at which a transformation occurs of austenite into something else has a profound effect on the metallographic structure produced, and mechanical properties are determined by the structure.

The transformation temperatures may be determined, for example, by noting changes of direction (arrests) in the heating and cooling

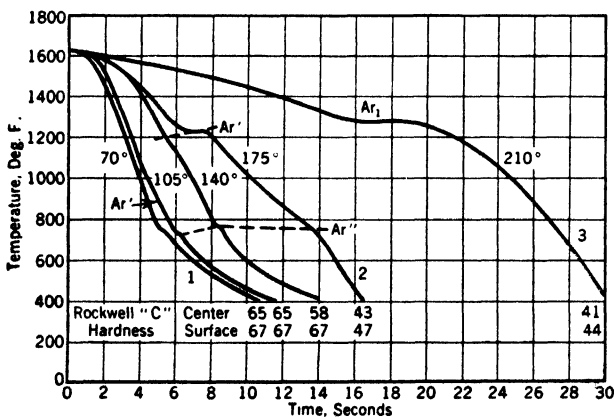


FIG. 6. Time-temperature cooling curves taken at the center of  $\frac{1}{2}$ -in.-diameter cylinders of 0.95% C steel when quenched from 1605° into still water, with the water at the temperatures shown. (French and Klopsch<sup>1</sup>)

curves of temperature against time (determined by suitable equipment) since heat is evolved on cooling, or absorbed on heating through a transformation. The volume changes likewise, and so a dilatometer can be used, and length plotted against temperature, to give the same information.

The arrests which supposedly would be obtained on *infinitely* slow heating or cooling and shown in the equilibrium diagram, are termed *Ae* (*e* for equilibrium). The temperatures of the *Ae* points are really averages of the *Ac* and *Ar* temperatures, obtained on very slow heating and cooling, since even under these conditions, they do not

exactly coincide. Under practical cooling conditions,  $Ar_1$  and  $Ar_2$  occur at lower temperatures than are indicated in the diagram.

Simple time-temperature cooling curves are shown in Fig. 6.

The "recalescent point"  $Ar_1$  can be considerably depressed even by fairly slow cooling. A plot of the amount of depression, at a fairly slow rate of cooling, against the carbon content, tends to follow the contour of the lower boundary  $GSE$  of the austenite field, as the upper curve in Fig. 7 shows. Although the temperatures are depressed, the points are still designated as  $Ar_1$ .

When a still faster rate of cooling is imposed, or as the austenite is made more sluggish, the depression increases and finally, as the second curve from the top in Fig. 7 shows, in steels of the C contents mostly used for heat treatment, a second recalescent point appears, and at a very much lower temperature.

The equilibrium diagram, Fig. 3, gave no hint whatever that this could occur! Figure 7 plots only the temperatures of the arrests; it does not show their intensity, but dilatometer curves or thermal-analysis curves at different rates of cooling give this information. There are volume changes attending the transformation, and so either thermal or dilation data will serve. Highly alloyed sluggish steels, studied at rates of cooling slow enough to be followed with fairly simple equipment, illustrate the principle. Figure 8 shows differential\* cooling curves obtained by Dejean<sup>2</sup> some

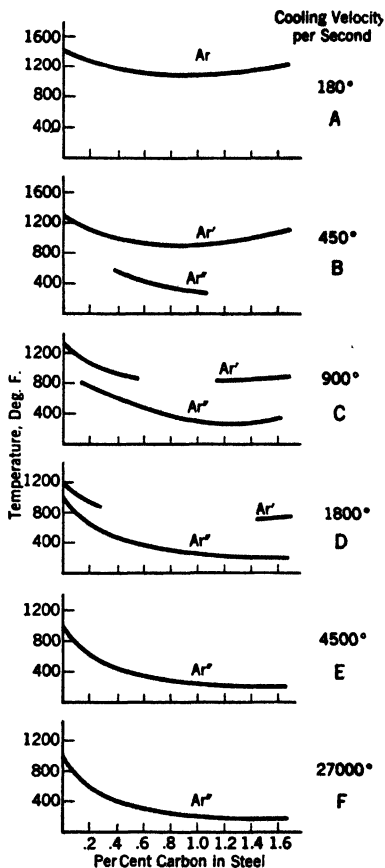


Fig. 7. Positions of  $Ar_1$ ,  $Ar'$ , and  $Ar''$  at different cooling rates. (After Esser, Eilender and Spenlé<sup>2</sup>)

\* See *ASM Handbook* for description of time-temperature, differential, and inverse-rate curves.

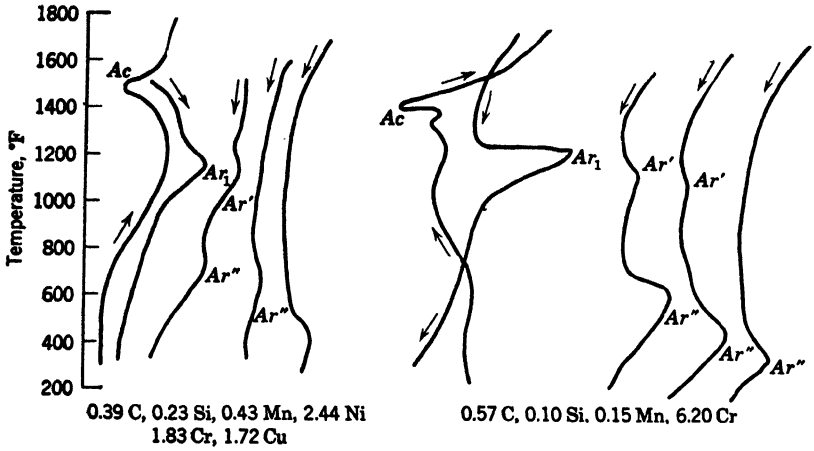


FIG. 8. Differential cooling curves. (Dejean <sup>3</sup>)

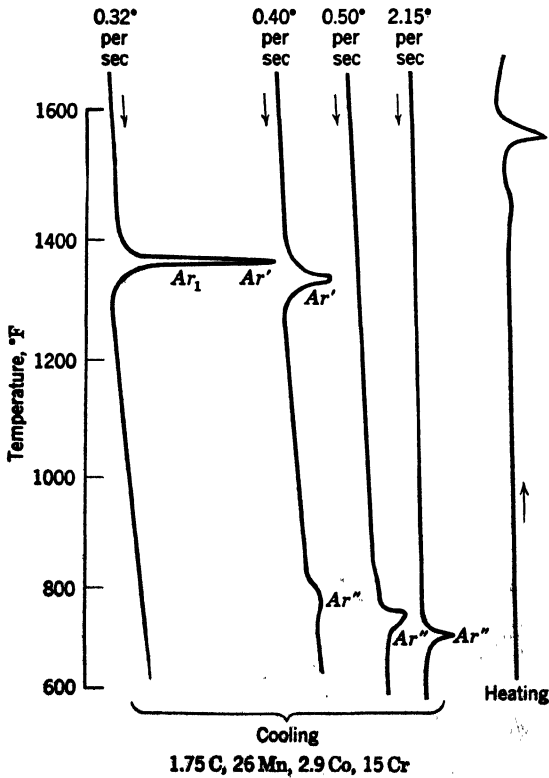


FIG. 9. Inverse-rate cooling curves. (Scott <sup>4</sup>)

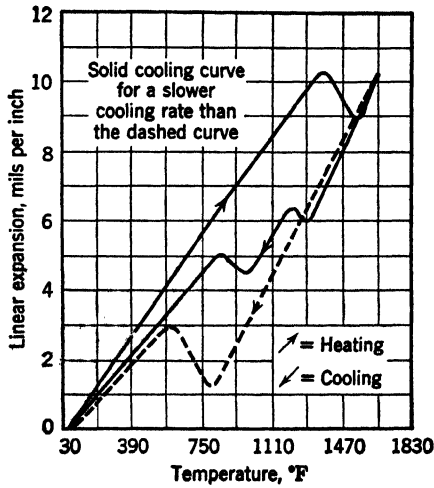


FIG. 10. Dilatometer curves for a 0.32 C, 0.69 Cr, 0.41% Mn steel. (Gillmor<sup>5</sup>)

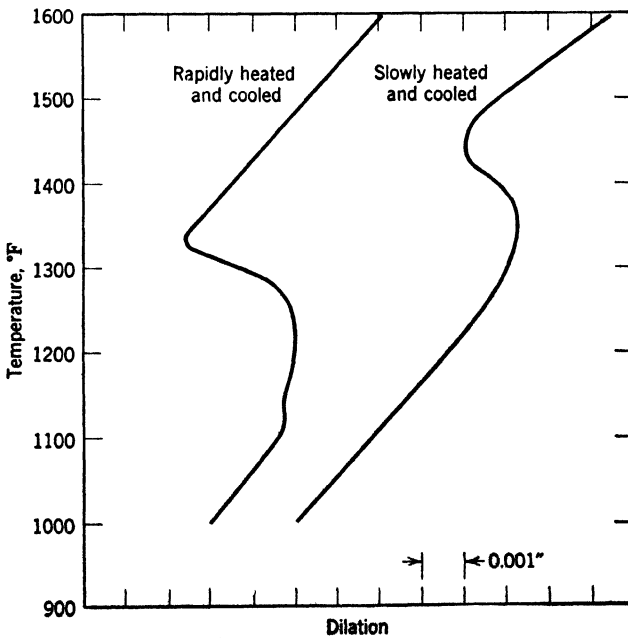


FIG. 11. Dilatometer cooling curves of a steel showing a split transformation in a rapid cycle but not in a slow cycle (0.15 C, 1.29 Mn, 0.29 Si, 0.014% Ti). (Battelle)

30 years ago. In these the effect is actually obtained by (1) increasing the temperature at which and the time for which austenite is held before cooling and (2) by using faster cooling rates. Figure 9, on an unusual very highly alloyed steel, shows the same behavior when cooling rate alone is varied.

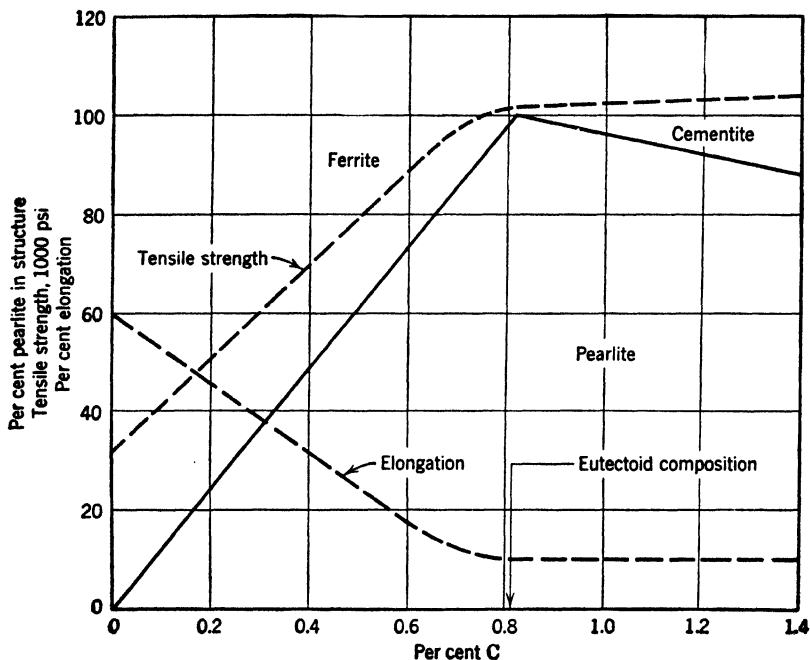


FIG. 12. Ferrite-pearlite-cementite diagram (solid lines), compared with tensile strength and elongation of annealed steels.

A dilatometer heating curve and cooling curves at two different rates for a steel of more usual composition are shown in Fig. 10, and dilatometer curves of a mild alloy steel in Fig. 11. This series of curves, Figs. 6-11, for various steels, shows that the behavior is general; that is, as the cooling rate increases and the second transformation appears, the upper transformation  $A_{r1}$  is not only lowered but also weakened. With still further increase in cooling rate  $A_{r1}$  gets still weaker, and the lower transformation stronger, until finally  $A_{r1}$  has disappeared and only the lower one is present.

*The Split Transformation.* The literature calls the appearance of two arrests on cooling a "split transformation." When the two

appear, the upper, normally called  $Ar_1$ , is then termed  $Ar'$  and the lower  $Ar''$ .

What this behavior really means is that *austenite is being retained* and that part of it transforms in one fashion, part in another, the  $Ar$  or  $Ar'$  transformation being to pearlite, the temperature being lower than the equilibrium diagram's slow-cooling condi-

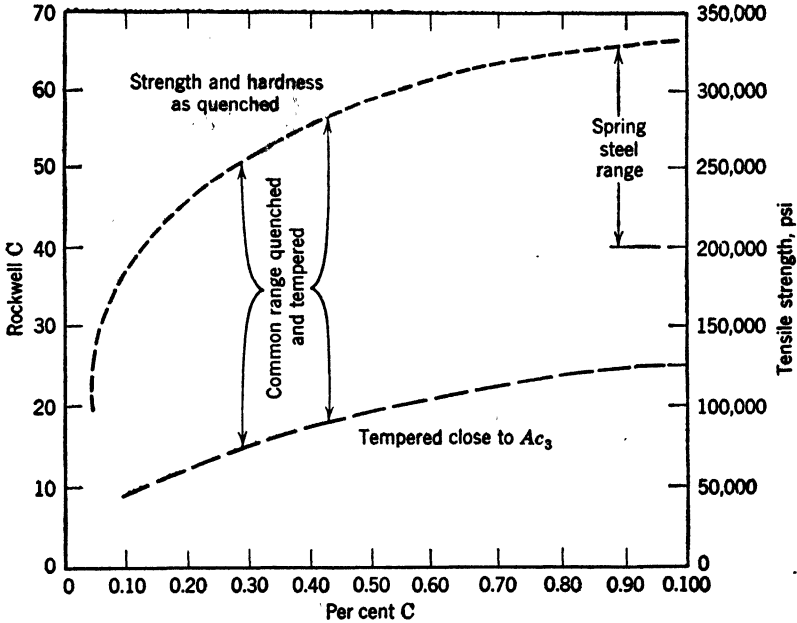


FIG. 13. Range of strength in quenched and tempered steels, depending on carbon content and tempering temperature.

tions call for, but the product still being pearlite, or an analogous product, upper bainite.  $Ar''$ , however, records the transformation of retained austenite not only at a much lower temperature, but also to entirely different products.

$Ar''$  is not indicated by the equilibrium diagram. This change is to *martensite*, often accompanied by "lower bainite." Both have needlelike, *acicular* structures.  $Ar''$ , when depressed to a sufficiently low temperature, becomes practically equivalent to  $M_s$ , a shorthand designation for the temperature of the *start* of *martensite* formation.

**Trigger Action.** If we revert to Fig. 7, the two middle plots show that the  $Ar'$  line disappears first at C contents near the eutectoid and that it is harder to retain austenite without having it go over to



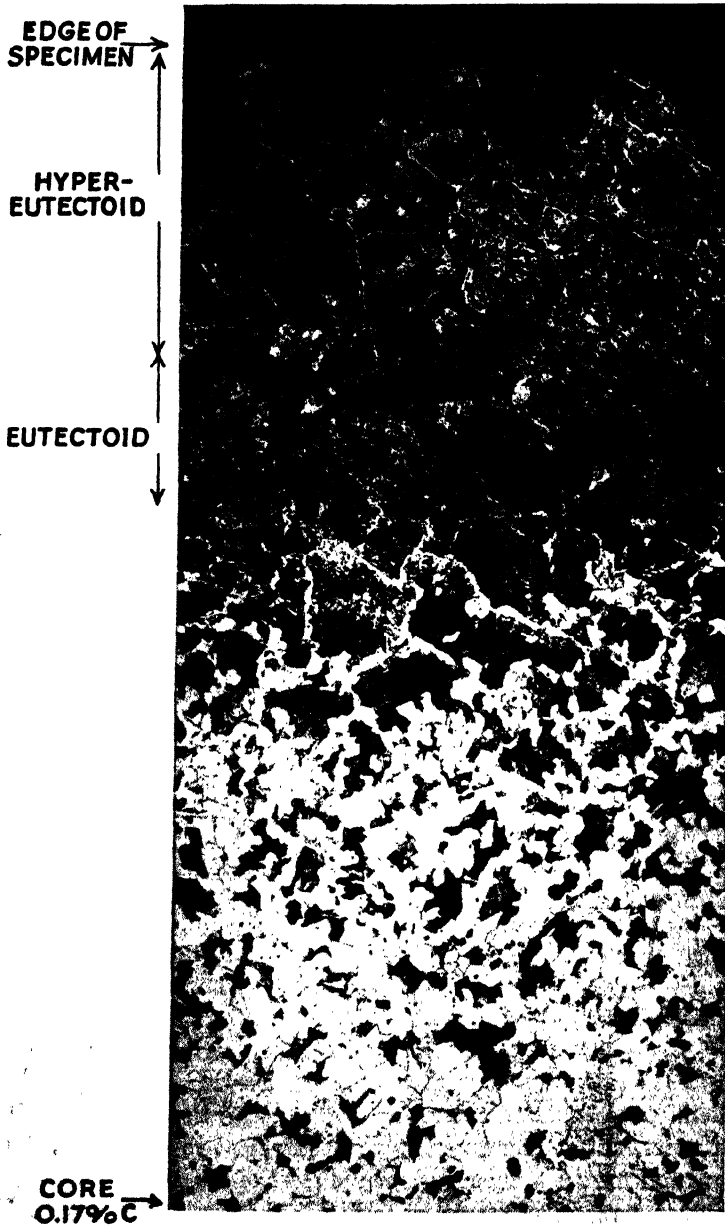


FIG. 14. Gradation of structure in carburized zone of a 0.17% C steel. Box-carburized 8 hr at 1700°. Slow-cooled. Structure normal, grain size 2 to 4.  $\times 75$ . (Battelle)

pearlite when the C is very low or very high. The reason is that the decomposition of austenite to give ferrite in hypo- and cementite in hypereutectoid steels wants to occur at higher temperatures (see the *GSE* line in Fig. 3), and thus the urge to transform is greater. If any ferrite or cementite is thrown out, its presence is usually considered to act as a sort of trigger to facilitate the pearlite transformation, since pearlite contains both, and either one might act as a nucleus. Cementite is now considered <sup>6</sup> to be the primary nucleating agent for pearlite.

**Strength versus Structure.** Thermal or dilatometric tests do not indicate *what* structures are being produced, merely that *some* transformation is going on. There are, however, material differences in the structures produced by transformation, according to the temperatures at which transformation occurs. The properties and behavior of the various structures naturally vary.

Strength varies with C content and with heat treatment. How great is the range of strength obtainable by adjusting both is brought out in Figs. 12 and 13. The range is from 35,000 to 350,000 psi tensile strength. The heavy lines of Fig. 12, for annealed steels, indicate the percentage of pearlite in the structure. The strength goes along with the percentage of pearlite. The ductility, as indicated by elongation, falls as the ferrite decreases.

Figure 13 shows the range in strength between the quenched (martensitic) condition and the fully tempered condition, for C steels of different C contents. Ductility, not shown, is inversely related to strength.

Figure 14 shows that the structure of slowly cooled steel changes with the C content. For a given C content, the structures, and hence the properties, depend on the temperature at which austenite transforms in cooling, and on the temperature to which martensite, or any other of the decomposition products of austenite, is reheated. Time is also a factor. A very vital matter is the temperature at which austenite transforms on cooling, since this governs the structure, and so we next examine the structures obtained at different temperature levels.

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## CHAPTER 3

### METALLOGRAPHIC STRUCTURES AND ASSOCIATED PROPERTIES

Definite rates of cooling produce definite structures, each with characteristic properties. The metallographic structure is an important clue to properties. Certain definitions relating to these topics are needed here.

**Micrography.** Microscopic examination of steel involves polishing the specimen to make it flat and smooth. *Nonmetallic inclusions* are visible without etching. Examination in the unetched condition appraises the steel as *clean* or *dirty*. *Magnafluxing*, discussed on p. 210, is a further aid. *Etching* is the application of some chemical reagent which attacks the different constituents of the structure differently and thus makes the structure visible. For appraisal of the structure revealed under the microscope, the etchant used needs to be known, as well as the magnification. Magnification is denoted by  $\times$ ; thus  $100\times$ , or  $\times 100$ , means that the linear dimension of the *micrograph* (photograph of the image produced in the microscope) is 100 times that of the part of the actual object shown. Thus the area is 10,000 times as great.

Linear magnification is ordinarily usefully made with the usual optical microscope up to about  $2000\times$ , sometimes higher with specially refined technique (Fig. 23c).

The *electron microscope* can give very much greater magnifications, and is being applied to some metallographic problems (Figs. 4 and 30).

**Homogenization.** When steel is heated for quenching or for normalizing or annealing, the way the center heats up is governed not only by the external conditions, but also by the size of the piece.

After the transformations of ferrite, pearlite, and carbide are completed by heating and everything is in solid solution in austenite (*austenitizing*), those localities originally in contact with carbide plates or particles are at first richer in C than localities more remote. Time or increased temperature is necessary to allow com-

plete *diffusion* of C and of any alloying elements present to produce *homogenization* of austenite. The degree of homogenization affects the behavior on cooling.

**Grain Growth and Grain Size.** As homogenization of austenite in respect to C and to any alloying elements present becomes more complete, there is a tendency toward *grain growth* (that is, a *coarsening* of the austenite grain).

The *grain size* of the austenite has a profound effect on hardenability and on some of the mechanical properties of the structures obtained either in normalizing or quenching. Certain additions to steel greatly raise the temperature at which *grain coarsening* occurs. This grain size is determined in various ways, and is usually described by comparison with ASTM Standards (p. 88). This refers to the prior austenitic grain size. In certain tool steels the prior austenitic grain size is apparent on fracture and comparison of the fracture with a set of steel standards. This *fracture grain size* is discussed on p. 371. These fine or coarse grain sizes are a different thing from the general term *fineness of structure*, which refers to the size and distribution of phases within a grain; whereas *structure* refers to the appearance under the microscope of the products of transformation of austenite rather than the prior austenite grain.

When steel is carburized to a hypereutectoid C level and slow-cooled, the pearlite grains may be surrounded by thin envelopes of carbide (this is called the *normal structure*), or the cementite may be present in patches (this is called *abnormal structure*). The origin of this terminology and figures illustrating the structure are given on p. 88.

Steel may be *decarburized* on the surface when heated, the C content at the surface being materially lower than it was originally. Decarburization is a serious handicap to heat treating and may be fatal to the life of a piece under repeated stress, because of the weakening of the surface.

### THERMAL BEHAVIOR

Thermal changes may merely produce reversible dimensional changes if there is no change in structure. When the structure is altered by transformation or by tempering, permanent changes appear. Even when no transformation occurs, the dimensional changes resulting from the cooling of a heated piece of steel set up *internal stresses* which may be retained after cooling is complete. The

stresses may exert a pull or a push; the former is a tensile, the latter a compressive, stress. They tend to produce warpage. The retained stresses within a piece must be in balance; a compressive stress at one location must be balanced by a tensile stress at another. Compressive stress at the surface is often useful; tensile stress at the surface usually harmful, as is discussed in Chapter 6. Cutting away part of a piece containing balanced stresses causes a redistribution of stress, and the piece warps.

Relief of internal stress is accomplished by heating into and slow cooling from the austenite range, that is, bringing about a complete transformation of the structure. Most of the stress can, however, be relieved by heating within the temperature range of existence of ferrite, even so low that no apparent change of structure results. This is a *stress-relief anneal*.

Another effect possible in the ferrite temperature range is *precipitation hardening*, discussed in Chapter 10. One effect of a type of precipitation hardening is *temper brittleness*, resulting from precipitation occurring during slow cooling after tempering. Still other low-temperature effects are *aging*, *quench aging*, and *strain aging*, also discussed in Chapter 10.

These factors all affect the result of heat treatment, in that we may be cooling austenites of different characteristics. In heating for tempering, the behavior will differ according to what transformation products of austenite have been formed on cooling.

**Tempering.** *Tempering* is done below the austenite range of temperatures. Tempering used to be called "drawing the temper," and so *drawing* and *draw temperature* are synonymous with tempering and tempering temperature; but, since there is the mechanical operation of cold drawing, tempering is the better term. Both time and temperature are factors in tempering.

**Pearlite.** An important structure in steel is that of pearlite. It consists of an aggregate of plates of ferrite interspersed with plates of iron carbide, in a layer or *lamellar* structure, or with a less stratified structure made up of the same components. It is a *two-phase system*, formed from single-phase austenite. With slow cooling, the composition of each grain of the pearlite aggregate is fixed; it contains a definite amount of cementite, corresponding to about 0.83% C in a plain C steel. Most alloying elements lower the percentage of C necessary to form a structure consisting entirely of pearlite.

Although the percentages of Fe and Fe<sub>3</sub>C in pearlite produced by slow and moderately slow cooling do not vary greatly, the faster

the cooling, the more layers of Fe and Fe<sub>3</sub>C are produced, and they are thinner layers in those steels whose pearlite has the layer construction.

At still more rapid cooling the pearlite grains do not have so well-defined a layer structure. The composition also becomes a bit lower in C in hypoeutectoid steels.

These shifts in form and composition come at the lower part of the *Ar* arrest, but in a gradual transition from the production of normal pearlite in the upper part of the arrest. Thermal or dilatometric tests do not indicate *what* structures are being produced, merely that *some* transformation is going on. There are, however, material differences in the structures produced by transformation, according to the temperatures at which transformation occurs. The properties and behavior of the various structures naturally vary.

**Isothermal Transformation.** The way to find out what structure corresponds to the transformation of austenite at a given temperature is to make the austenite transform at that one temperature only. This is easier said than done, but within limits it can be done.

It should be recognized that, when we do it, we do it for just what information it will give, and that we cannot expect it to give accurate information for conditions of continuous cooling, when the austenite is never halted at one selected temperature. In practical heat treatment, except for "austempering" and "martempering," (which are discussed later), cooling is continuous. The cooling rate through the section will vary with the size of the piece and with the speed with which heat is abstracted from the surface. At slow-cooling rates there may be time for the transformation to occur so nearly at one temperature that the resulting structure will be fairly representative of that temperature. At an extremely fast-cooling rate austenite may all be retained down to a desired temperature and then held or cooled further, as we wish, but to accomplish this the piece must be very small so that the center cools practically as fast as the outside, or else the austenite must be made so sluggish, by alloying, that it does not transform at the relatively slow-cooling rate at the center of a larger piece.

When neither of these two precautions is applied, as in the very common case of a large section of C or low-alloy steel, a great variety of cooling rates exists from surface to center, and the time each point at varying distances from the surface remains at a given temperature during cooling will vary widely; and, since the steel is not a sluggish one, it responds differently to the different cooling

rates. Hence it is possible to obtain, at different depths in one piece, all the structures characteristic of transformation at different temperature levels. Thus, in a "slack-quenched" piece, where the center is not cooled at a sufficiently rapid rate to escape transformation at a temperature above that for the formation of the martensite

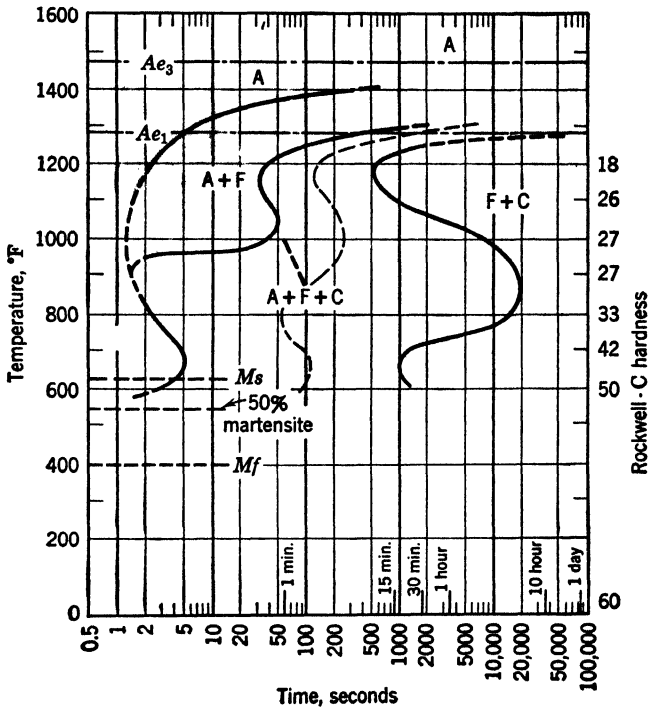


FIG. 15. A time-temperature transformation or S curve.<sup>1</sup> N.E. 9442, 0.38 C, 1.08 Mn, 0.70 Si, 0.40 Cr, 0.34 Ni, 0.11 Mo, 0.03% Zr, austenitized at 1575°F; grain size, 10-11.

we ordinarily seek to produce by quenching, there will inevitably be a gradation in structure from outside to inside.

So, to study what goes on at one temperature, very tiny pieces are used. They are heated to make them austenitic, then plunged into a molten salt or molten lead or other bath, and held at the temperature it is desired to investigate. After being held for seconds, minutes, or hours, the specimen is quenched in cold brine, in order to transform to martensite any austenite not previously transformed at the holding temperature in the time allowed. The structure is then examined under the microscope. Until the time at which trans-



formation starts at the holding temperature, the quenched specimen is *all* martensite, or martensite plus austenite, either case corresponding to *no* previously transformed austenite. At increasing holding times, part of the austenite is transformed, and part retained, the percentage transformed increasing until, with enough time at holding temperature, *all* the austenite is transformed, *no* martensite is formed by the quench, and the structure is characteristic of the transformation temperature level. The hardness of this fully transformed specimen is that characteristic of the structure. Unfortunately, the tiny specimens are too small for easy determination of mechanical properties other than hardness.

Figure 15, from the Atlas of Isothermal Transformation,<sup>1</sup> with the temperature range for martensite formation added, shows a "time-temperature transformation" (*TTT*) plot for N.E. 9442. Note that time is plotted on a logarithmic scale. Look first at the right-hand curve, the ending curve, for complete transformation of austenite. It has the shape of an S, and these isothermal plots are usually termed S curves, although not all have that contour. Note the hardness figures at the right margin, for the completely transformed specimens. As the reaction temperature drops, the hardness increases.

**Products Produced at Different Temperatures.** The dotted curve to the left of the ending curve is for the transformation of 50% of the original austenite. The S-shaped curve next toward the left is for the start of the separation of carbide. Above 900° there is a branch of the curve showing the start of formation of proeutectoid ferrite, since this is a hypoeutectoid steel. This branch would be absent in a eutectoid steel. In the plot *A* is austenite; *F*, ferrite; *C* carbide.

At about 1150° there is maximum urge for transformation to pearlite; its formation starts earlier and is completed sooner than at higher or lower temperatures. This is the "pearlite nose" of the curve. At about 900° ferrite ceases to be thrown out separately, and ferrite plus carbide come out together, but not in the pearlite ratio. Here "upper bainite" is forming. It has, in this steel, a great urge to form, starts to form quickly (importantly, it does take over a second for it to start to form), but, since the steel is rather stiff at this temperature, it takes a long while, some 5½ hr, for the transformation to be completed, whereas, at 1150°, 10 min sufficed. This 900° peak in the left-hand curve is the "bainite chin" of the S curve.

At the still lower temperature of  $700^{\circ}$ , after 5 sec, "lower bainite" starts to form, and in some 20 min all the austenite transforms to this structure (an acicular one) with a Rockwell cone hardness of 42.

At about  $610^{\circ}$  a different condition exists. Martensite starts to form at this temperature. In *continuous* cooling, martensite keeps on forming as the temperature drops, the steel then becoming all martensitic at about  $400^{\circ}$ . However, at intermediate temperatures, two modes of transformation are possible, either to martensite or to lower bainite. Consider  $600^{\circ}$ . In quenching into a molten bath at  $600^{\circ}$ , a little martensite will form, as a result of rapid cooling from  $610^{\circ}$  to  $600^{\circ}$ . If the piece is held at  $600^{\circ}$  for some 20 min, the remaining austenite transforms to lower bainite—at constant temperature—with a resulting hardness of 50 Rockwell cone. If, on the other hand, rapid cooling is continued down to  $400^{\circ}$ , the remaining austenite transforms to martensite, just as the first few per cent did, and the resulting hardness of the *all*-martensitic structure is 60 Rockwell cone.

**Extent of Information from Isothermal Studies.** The isothermal method of study thus can be made to give much more information than is obtained from the cooling curves shown in the previous chapter. Those told only that two different changes (the split transformation) were going on during continuous cooling, but did not allow checking up on the structure produced at any one temperature. The isothermal technique tells, for each temperature, what transformation occurs, how fast it starts, how long it takes to complete it, what structure is produced, and how hard that particular structure is. This is all useful information, and information from S curves is used herein to explain many phenomena of heat treatment. However, the isothermal S curves do *not* give direct information applicable to all conditions of continuous cooling. As is brought out later, technique can be modified to produce a type of S curve giving much more directly applicable information.

## STRUCTURES OF THE PRODUCTS

Having described the isothermal method of study, we may now turn to the structures characteristic of a given temperature level. It may be understood that Fig. 15 is merely one sample S curve, steels of other compositions and other C contents have very different

S curves, and for each steel the contour of the curve, the temperature and time locations of the pearlite noses and bainite chins will vary. The structures and hardnesses produced at the same temperatures will vary from steel to steel.

However, there is a general pattern running through all steels. As the transformation temperature at which pearlite is formed decreases, the pearlite becomes finer and harder. At still lower temperature upper bainite, a still harder nonlamellar product, forms. Still lower in temperature the structure becomes progressively finer and acicular (lower bainite), until around the  $M_s$  level, it greatly resembles martensite, and approaches but does not reach the hardness of martensite. It is emphasized later that martensite forms *only* on a falling temperature, but lower bainite can progressively form at a *constant* temperature.

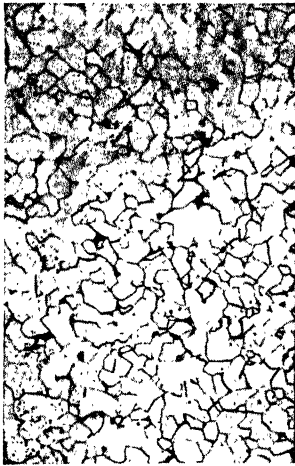
**Austenite.** Since austenite in C and low-alloy steels is not stable at room temperature, it is necessary to use a highly alloyed steel,



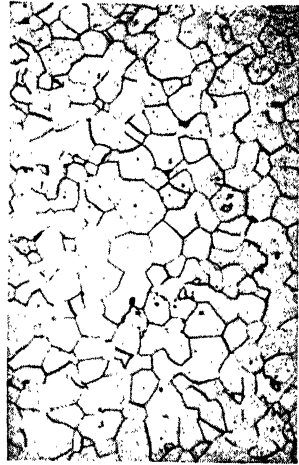
FIG. 16. Rolled manganese steel water quenched from  $1925^{\circ}$ .  $\times 100$ . (Battelle)  
1.36 C, 0.36 Si, 14.20% Mn.

Fig. 16, to show the austenitic structure. The bands and wrinkles across the crystals are due to deformations from rolling, not to local composition differences. In spite of the high-C content of this austenite, no carbides are visible, the C is all in solution in the austenite. Austenite consists of grains, each one like the other; it is a single phase.

**Ferrite.** In transforming to ferrite, on cooling, C-free austenite, that is, pure iron, breaks down into smaller crystals; in other words, the structure is "refined," but the size of the ferrite crystals depends on the time spent at  $A_r'$ ; fairly rapid cooling (Fig. 17, left) produces fine-grained ferrite, very slow cooling (Fig. 17 right), coarser and softer ferrite. Figure 17 is not of pure Fe, but of an 0.07% C steel, and a few dark pearlite particles, whose structure is not resolved at this magnification, appear.



(a) Normalized, 1850°, 15 min, air.



(b) Box-annealed, 1450°; heated and cooled at 50° per hr.

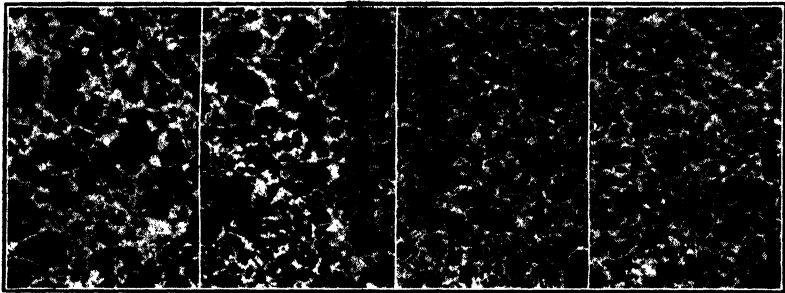
Fig. 17. Structure of 0.07% C steel.  $\times 100$ . (Battelle)

In a hypoeutectoid steel in which ferrite is first thrown out on cooling and the austenite enriched in C until it gets down to eutectoid composition, as in the upper part of Fig. 15, the ferrite accumulates as a network around the remaining austenite, which then transforms to pearlite.

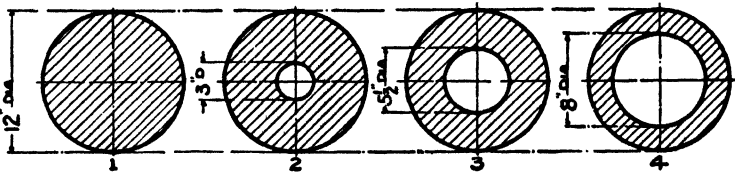
The thickness of the ferrite network and the size of the pearlite patches decrease as the temperature of transformation decreases, that is, as the cooling rate increases. This is shown in Fig. 18, where, going from left to right, the cooling rate in air is increased as the ratio of surface to volume is increased by boring out larger inside diameters.

At the relatively slow-cooling rates of Figs. 17 and 18 the outside and center cool at not too greatly different rates, and the gross structures are characteristic of transformation at approximately constant temperature.

**Pearlite.** But to examine the structure of pearlite itself, produced at different temperature levels, resorting to the isothermal technique brings out the important facts. Figures 19a and b show how much finer and harder is the pearlite produced at 1225° than at 1275° in a slightly hypereutectoid steel. (Variations in the apparent spacing of the pearlite and cementite layers in any one micrograph of pearlite



× 100.



Ratio Volume to Surface	8:1	2.25:1	1.6:1	1:1
Tensile	96,500	97,000	97,500	97,000
Yield	55,000	56,000	59,000	60,000
Elong. %	21	23	25.5	25
R.A. %	34	36	42	40.5

Steel of 0.34 C, 0.29 Si, 0.58 Mn, 0.034 P, 0.04% S.

Fig. 18. Effect of boring on mechanical properties and structure of normalized forgings of the same steel. (Fry<sup>7</sup>)

depend largely on the angle the surface shown makes with individual pearlite grains.)

**Upper Bainite.** As the temperature of transformation drops just below the pearlite nose of the S curve, the ferrite separation above it, in a hypoeutectoid steel, no longer occurs, and the layer structure of the pearlite is less evident. At low magnification, Fig. 20, in a specimen of slightly hypoeutectoid C steel, etched with nitric acid and alcohol, the structure appears as a confused mass with dark and light etching locations. This structure used to be called primary troostite, nodular troostite, or troostite of quenching, and was then considered as a rather individual type of structure. With higher magnification and good resolution, as in Figs. 21 and 22, it becomes

plain that the structure is still made up of ferrite plus carbide. Figs. 20-22 are for a continuously cooled, not an isothermal, specimen.

The temperature level makes a big difference in the arrangement of the constituents. This is true even as to the crystal shape of a



FIG. 19a. Coarse pearlite produced in an SAE 3100 series steel containing 0.91% C, isothermally transformed at 1275°F. Picral etch  $\times 1200$ . Hardness, 11 Rockwell C. (Elsa, Battelle)

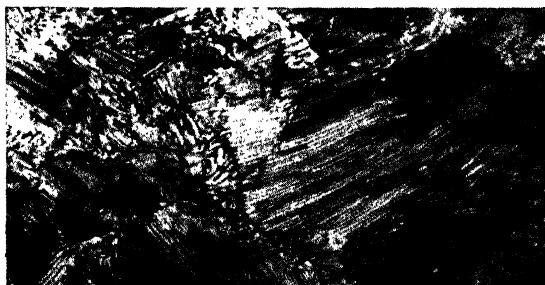


FIG. 19b. Fine pearlite produced in an SAE 3100 series steel containing 0.91% C, isothermally transformed at 1225°F. Picral etch  $\times 1200$ . Hardness, 26 Rockwell C. (Elsa, Battelle)

single constituent. Going back to a higher temperature level, one at which, in hypoeutectoid steels, proeutectoid ferrite is being thrown out, consider Fig. 23. Here N.E. 8740, which has the upper part of its S curve somewhat like that in Fig. 15, was used, plunged in a bath held at 1250°, for a time just short of that at which pearlite would form, while ferrite was separating, and then brine-quenched. This made the background martensite; the white patches are ferrite. This ferrite separated in blocky form.

But in the high Ni steel of Fig. 24, the temperature of  $A_{r'}$  is much lower, and the time the steel can be held at 950° without separa-

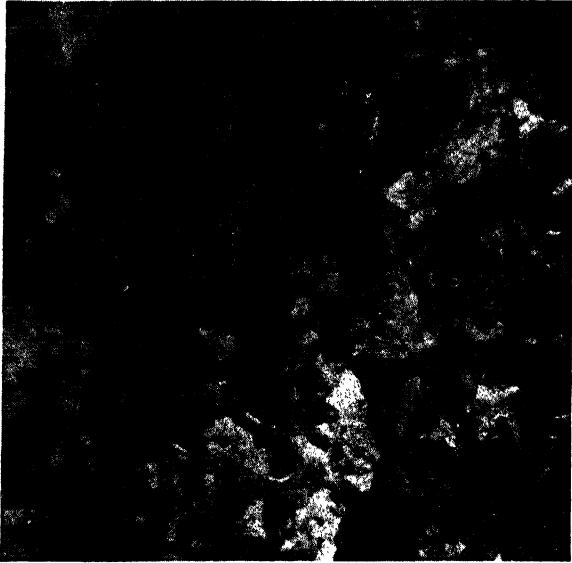


FIG. 20

× 200.



FIG. 21

× 1000.

FIGS. 20 and 21. Primary troostite at two magnifications. (Lucas<sup>1</sup>)

tion of pearlite, very long. At this lower temperature level, the separated ferrite is distinctly acicular.

**The X Constituent.** Some constituents can be shown up in isothermal studies, whose constitution is not well enough understood to be named. Such is the "X constituent" produced in a NiCr steel by the isothermal treatment described in the legend of Fig. 25.

Going now to the formation of bainite, we have mentioned "upper" bainite and "lower" bainite. They are named from the location of



× 4000.

Fig. 22. Primary troostite at still higher magnification. (Lucas<sup>1</sup>)

their formation temperatures on the "bainite chin" of the S curve for the particular steel in question.

Upper bainite produced isothermally is shown in Fig. 26. A case of its production on continuous cooling is shown in Fig. 28. It consists, in a hypoeutectoid steel, of carbide particles arranged in some degree of order, in a ferrite matrix. In a given steel, it is appreciably harder than the pearlitic structures produced at higher temperatures. Lower bainite is pictured in Fig. 27.

**"Troostite."** In slack quenching, upper bainite is often formed at the boundaries of austenite crystals, so that, after quenching, the structure shows both martensite and upper bainite, the latter in characteristic feathery form. This feathery boundary constituent



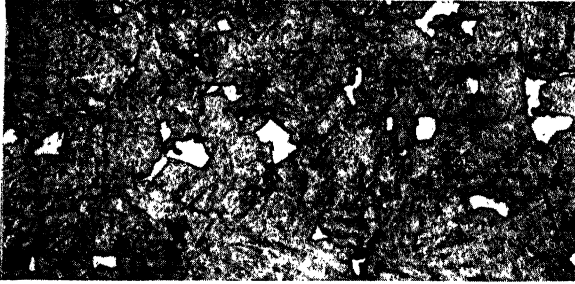


FIG. 23. Block ferrite in a matrix of martensite produced in N.E. 8740 steel isothermally transformed at 1250°F for 150 sec, followed by a water quench. Picral etch  $\times 500$ . (Elsa, Battelle)



FIG. 24. Acicular ferrite in a matrix of martensite produced by isothermal transformation at 950°F for 15 hr followed by a water quench. Picral etch  $\times 500$ . Steel contains: 0.34 C, 0.71 Mn, 0.22 Si, 2.52 Ni, 0.14 Cr, 0.42 Mo, 0.02% V. (Elsa, Battelle)

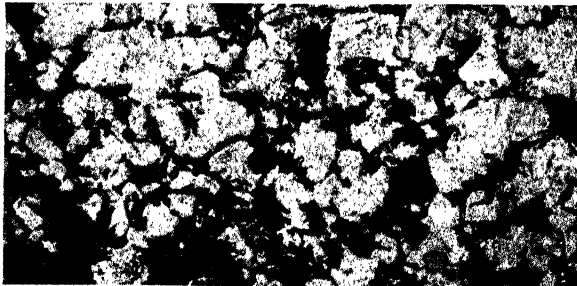


FIG. 25. X constituent in a matrix of martensite produced in SAE 3250 steel isothermally transformed 3000 sec at 990°F, followed by a water quench. Picral etch  $\times 500$ . This rather uncommon structure may be produced by isothermal treatment between the pearlite nose and the bainite chin of the S curve of a steel in which the nose and chin are distinct and quite far separated. (Elsa, Battelle)

was formerly termed "primary troostite" or "troostite of quenching" and was thought of as an entity. Figure 28 shows this duplex structure. If the first transformation occurs at a lower temperature,



FIG. 26. High-temperature bainite produced by isothermal transformation at 900°F for 10,000 sec. 0.30 C, 0.49 Mn, 0.41 Si, 0.51 Cr, 0.41% Mo. Picral etch  $\times 1200$ . (Elsa, Battelle)

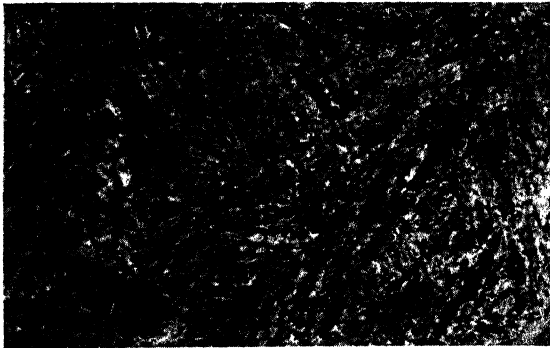


FIG. 27. Lower bainite produced in an SAE 3100 series steel containing 0.91% C isothermally transformed at 500°F. Picral etch  $\times 1200$ . (Elsa, Battelle)

lower bainite may be produced as shown in Figs. 27 and 29. These figures are explained in their legends.

**Lower Bainite Resembles Martensite.** Figure 29 shows lower bainite, isothermally produced. Note its similarity in appearance to martensite and partly tempered martensite, Figs. 31-33. Cuckon<sup>10</sup> suggests that the two can be differentiated by use of the electron microscope, or by special tricks with the light microscope. As the transformation temperature is decreased, the carbide arrangement tends more and more to the acicular, until, in "lower" bainite, the structure of Figs. 29 and 30 is produced.

Bainite is defined by Zener<sup>2</sup> as ferrite supersaturated with C, differing from alpha martensite (also defined as ferrite supersaturated with C), in that in bainite the C atoms are in a random arrangement, whereas in alpha martensite, they are in a more definitely ordered arrangement. The internal stress accompanying the



× 250.



× 250.

FIG. 28a. Heavy aqua regia etch.

FIG. 28b. Light aqua regia etch.

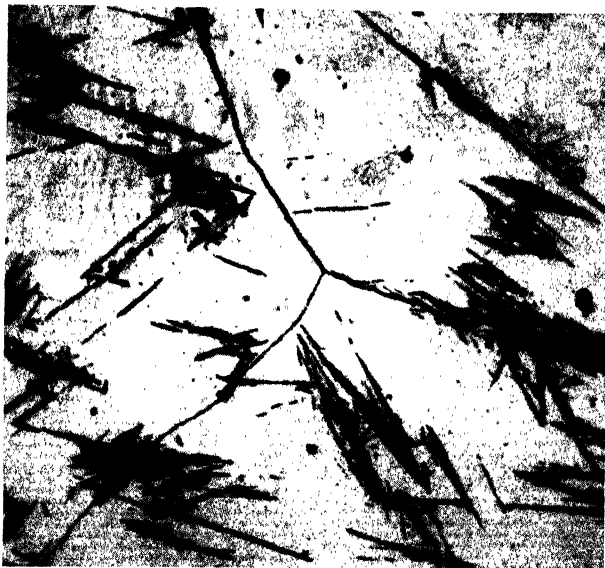
Upper bainite, formed at grain boundaries during early stages of cooling of a rapidly cooled casting of 0.45 C, 11.0 Cr, 1.3 Si, 0.8 Mn, 0.1% Ni alloy steel. The grains transformed to martensite at a later stage. Hardness, 600 Brinell. At much higher magnifications the bainite would appear similar to that in Figs. 21 and 22. Similar structures may be obtained in the slack quenching of a less sluggish steel. (Brasunas, Battelle)

ordered arrangement in alpha martensite is considered to be absent in the case of bainite.

That the acicular structure of Fig. 30a is due to the arrangement of carbide particles is brought out in Fig. 30b. The needles in Fig. 30a look like individual crystals, but the electron micrograph, Fig. 30b, shows that they are made up of smaller carbide particles, thrown out in certain crystallographic planes of the austenite crystal, just as is shown in Fig. 28.

Thus many different structures may be produced by different modes of deposition of ferrite and carbide, thrown out of austenite held at different temperature levels for different times.

Austenite retained down to the  $M_s$  temperature, and cooled through the  $M_s$ - $M_f$  range fast enough to escape transformation to bainite, transforms to martensite. With suitable steels and cooling rates *all* the austenite changes to martensite, but under some cooling conditions many high-C or high-alloy steels retain austenite.



× 250.

Fig. 29. Aqua regia etch. Lower bainite in a rapidly cooled casting of 0.47 C, 16.0 Cr, 1.3 Si, 0.8 Mn, 0.1 Ni, 0.03% N alloy steel. This is a still more sluggish steel than that of Fig. 28. Lower bainite was formed at a lower temperature than in Fig. 28. It separated not only at grain boundaries, but also along certain crystallographic planes of the austenite crystal. Some of the austenite later transformed to martensite, the light-gray patches in the white untransformed austenite matrix. (Brasunas, Battelle)

As is explained later, this residual austenite may be coaxed to transform, but the conditions attending its transformation are a bit complex.

**Martensite.** Heat treatment by quenching and tempering revolves about the retention of austenite to such a low temperature that martensite is formed.

Consider a steel quenched to 100% martensite, with resulting acicular structure as shown in Fig. 31. There is marked similarity between this and the acicular low-temperature bainite shown in Fig. 27, but the martensite is harder and is indeed a different product.

Freshly quenched martensite of sufficient C content has a *tetragonal* structure called *alpha martensite*. It contains ferrite and C. The C is probably there as C, or possibly as very finely divided  $\text{Fe}_3\text{C}$ . On slight heating (tempering) the tetragonal structure changes to regulation body-centered cubic ferrite. In this condi-

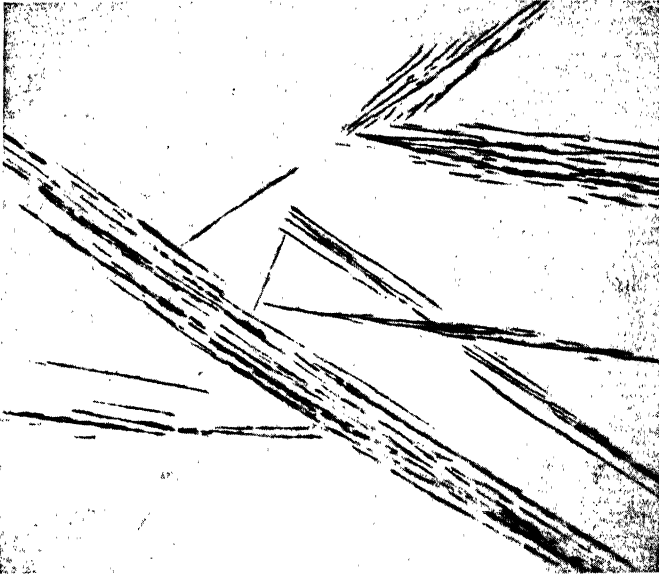


FIG. 30a. Bainite in a matrix of austenite produced by isothermal transformation at 590°F for 25 hr. Picral etch  $\times 1200$ . Steel contains: 1.2 C, 1.5 Mn, 2.25 Cr, 0.40% Mo. Austenite is easily retained in this steel since  $M_s$  is low. The small amount of martensite produced in cooling from 590° is not visible with this etch. (Elsea, Battelle)

tion the martensite is termed *beta martensite*. The C, in alpha martensite, on slight tempering, combines to form  $\text{Fe}_3\text{C}$  or, if it is considered that the  $\text{Fe}_3\text{C}$  was originally present on quenching, it agglomerates somewhat. Low-C martensite may form originally as beta or shift to beta at once.

Although the formation and/or separation of  $\text{Fe}_3\text{C}$  is usually considered to follow closely upon the alpha to beta martensite change, there is some evidence<sup>8</sup> that the first separation of a compound is of  $\text{Fe}_3\text{N}$ , followed, at a somewhat higher tempering temperature, by  $\text{Fe}_3\text{C}$ . At any rate, as tempering progresses,  $\text{Fe}_3\text{C}$  is formed.

**Structure of Martensite.** The crystallographer's explanation of the tetragonal structure of alpha martensite is that in the face-centered cubic structure of austenite the Fe and C atoms are so arranged that the structure could also be considered as made up of rectangular prisms, the height being considerably greater than the length of the sides of the square base. In transforming to ferrite, pure iron shifts from this austenitic structure to the body-centered



Fig. 30b. Electron micrograph of the same specimen  $\times 10,000$ . Formvar replica, shadow-cast with gold. (Schwartz, Battelle)

cubic structure without difficulty. But when C is present in the austenite and the transformation of austenite is inhibited by rapid cooling so that it does not occur until the low temperature of martensite formation is reached, the transformation, which can be looked at as the shoving down of the top of the rectangular prism until it becomes a cube, stops or slows down before the top is completely shoved down, the presence of trapped C in an orderly arrangement being held responsible.

Thus, in alpha martensite, the structure is a rectangular prism with its height very slightly greater than the sides of the square base, but enough greater to be definitely provable by X-ray crystallographic methods. Long standing or gentle heating permits this crystal to throw out the trapped C (into random arrangement, thus

relieving internal stress which accompanied the prior ordered arrangement) and complete its shrinking down to the perfect cubical form of the ferrite in beta martensite.

The alpha martensite is termed tetragonal rather than prismatic, when thought is fixed on the position of the C atoms. In austenite a C atom is considered to be bound to Fe atoms so as to form an octahedral inner structure; in alpha martensite, so as to form a tetrahedral inner structure. In beta martensite the C is considered either to be temporarily free and then to form  $\text{Fe}_3\text{C}$ , or at once to form  $\text{Fe}_3\text{C}$ .

These crystallographic explanations seem to have evolved from a suggestion by Bain for the mechanism of the martensite transformation. The alpha-beta martensite change is described by Houdremont.<sup>4</sup> Recent discussions are given by Lipson and Parker,<sup>5</sup> by Antia, Fletcher, and Cohen;<sup>6</sup> and Zener.<sup>2</sup> These crystallographic explanations are of little direct concern to the heat treater but tend to satisfy one's natural longing for explanations of the mechanisms concerned in the transformations.

**Tempering of Martensite.** Slightly tempered beta martensite, made up of ferrite with very tiny and uniformly dispersed carbide particles, is still very hard and brittle, though wear-resistant. With only slight tempering, it is used, mainly for its wear-resistant properties, as the thin surface of carburized pieces. For most constructional purposes a piece originally all martensite, or with a thick martensitic shell, is tempered high enough to be at least slightly softened and materially toughened. This softening and toughening result from the agglomeration of carbide, small particles coalescing to form larger ones and thereby creating wider paths in the ferrite matrix.

The maze has become less tortuous. Tiny hard brittle particles in a soft matrix reinforce the matrix and harden it. If the hard particles are too brittle and too close together, the whole mass behaves in brittle fashion. When the soft matrix has wider less interrupted paths, its influence is increased, and some measure of toughness produced.

**"Secondary Troostite."** A still hard but fairly tough structure is produced by tempering the martensite of Fig. 31 at  $750^\circ$ , giving the structure of Fig. 32. Because of agglomeration of carbide, the acicular structure is even more readily visible than it was in Fig. 31. It used to be considered that martensite tempered as in Fig. 32 rated a special name, "troostite," or, to differentiate it from the "primary

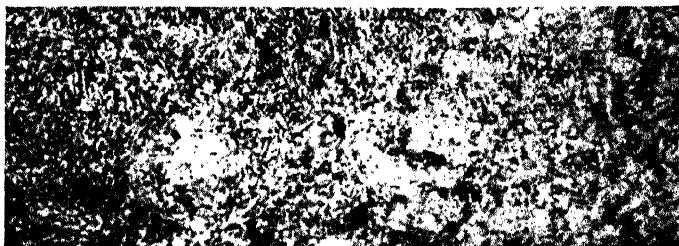


FIG. 31. As water-quenched from 1650°, 60 R.C.

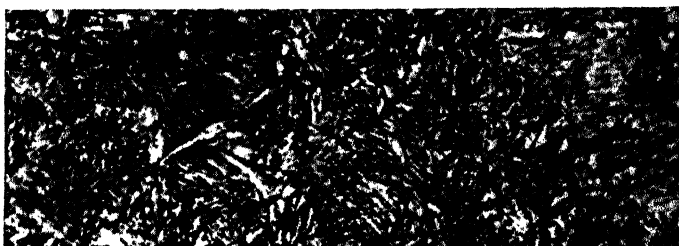


FIG. 32. Tempered at 750°, 45 R.C.

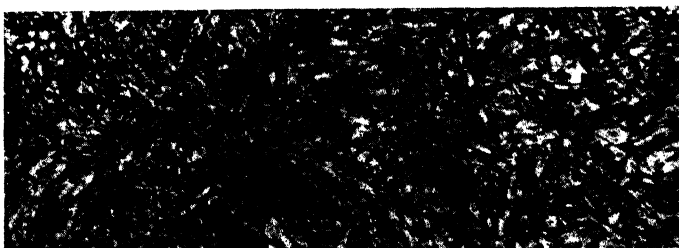


FIG. 33. Tempered at 1120°, 34 R.C.

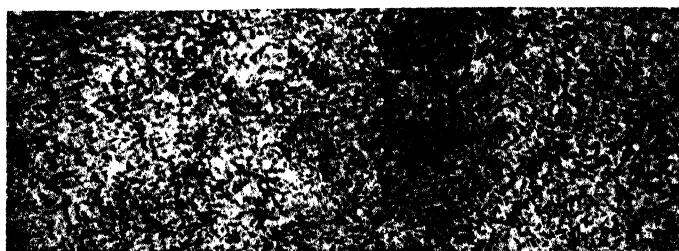


FIG. 34. Tempered at 1250°, 25 R.C.

FIGS. 31-34. Structures of martensite before and after tempering 1 hr at increasing temperatures. Analysis: 0.42 C, 0.30 Si, 0.89 Mn, 0.58 Ni, 0.52 Cr, 0.24% Mo.  $\times 1200$ , Picral etch. (Elsa, Battelle)



troostite" of Fig. 27, "secondary troostite." In early days, before the isothermal method of study, it was difficult to differentiate between even upper bainite and martensite. A slightly slack-quenched piece, mostly martensite, but with some upper bainite, tempered as in Fig. 32, etched with nital and looked at with a microscope of less than present-day resolving power, had dark etching patches where the upper bainite had tempered faster than the martensite, showing that two different structures were originally present. Since the previous presence of bainite was not recognized, and it was supposed that the dark patches had originally been martensite, it was thought that "secondary troostite" came from martensite.

"**Sorbite.**" With severe tempering, as in Fig. 34, considerable carbide agglomeration occurs, and in the older nomenclature the structure of Fig. 34 was called "sorbite." To the structure produced by intermediate tempering, as in Fig. 33, the term "troosto-sorbite" was given. These terms still persist in some current literature, but are being given up because, as Figs. 32-34 show, the tempering of martensite at increasing temperatures produces a *gradual* modification of the structure, without abrupt changes. These structures are now all termed "tempered martensite."

The structures, austenite, martensite, bainite, troostite, and sorbite were named in honor of Roberts-Austen, Martens, Bain, Troost, and Sorby, metallographers who were pioneers in the study of steel.

✓ **Spheroidization.** If the tempering of Fig. 34 had been carried on much longer, still further agglomeration of carbide would have occurred, the carbides finally becoming rounded, that is, "spheroidized."

Spheroidization of carbide, by long heating just below  $A_{c1}$ , goes on, whether the starting material is tempered martensite, or pearlite produced by normalizing or annealing. Figures 35-37 show structures in an annealed large forging of 0.45 C, 0.57% Mn steel, as spheroidization progresses.

TABLE 2

Fig.	Treatment	Tensile Strength, Psi	Yield Strength, Psi	Elongation, % in 2%	Reduction of Area, %
35	1475° 1 hr furnace-cooled	84,750	40,250	19.5	28
36	1250° 48 hr	72,250	35,000	24.5	34
37	1250° 120 hr (72 hr after the first 48-hr treatment)	69,500	33,250	30.0	52



FIG. 35. First treatment.  $\times 1000$ . (MacPherran and Harper<sup>6</sup>)



FIG. 36. Second treatment.  $\times 1000$ . (MacPherran and Harper)

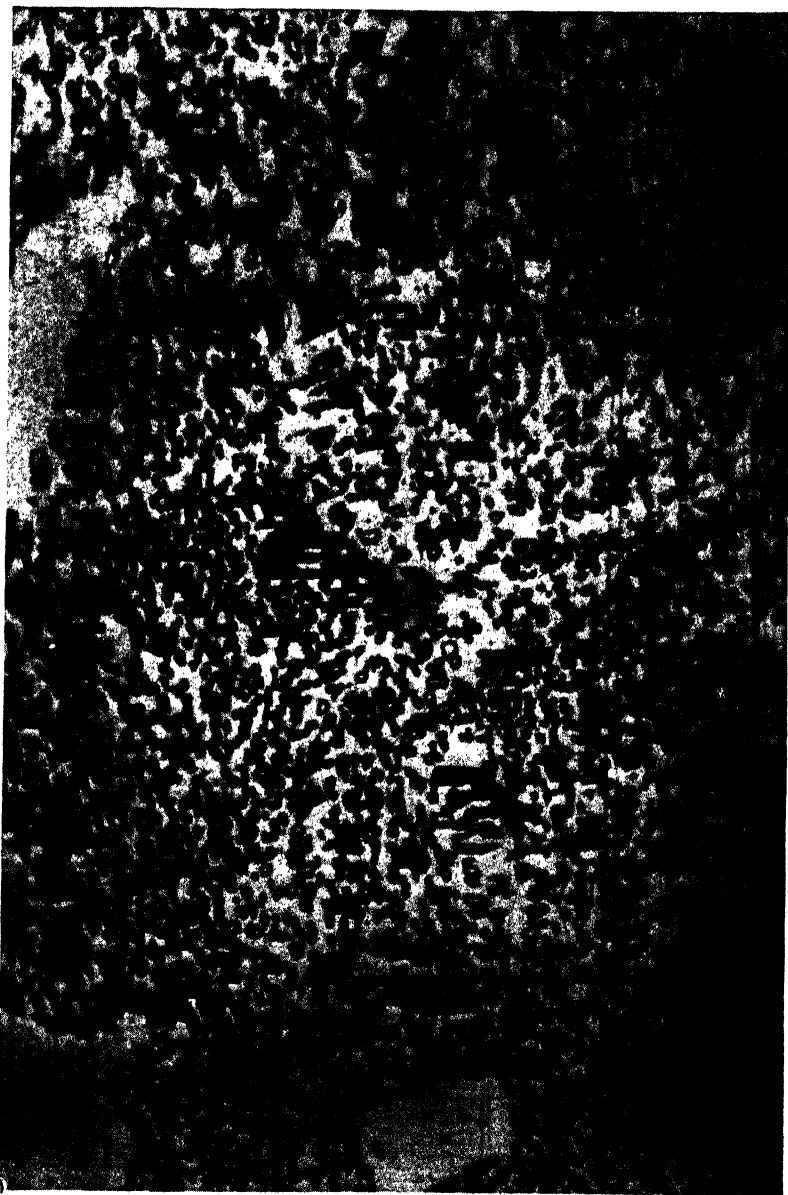


FIG. 37. Third treatment.  $\times 1000$ . (MacPherran and Harper)

The softening and toughening produced by spheroidization and consequent opening up of less obstructed ferrite paths through what was formerly pearlite are evidenced by the tensile data on radial specimens (Table 2).

**Preparation for Hardening.** An all-austenite structure cannot be preserved at room temperature in C steels. To accomplish this, a large amount of one or more suitable alloying elements must be

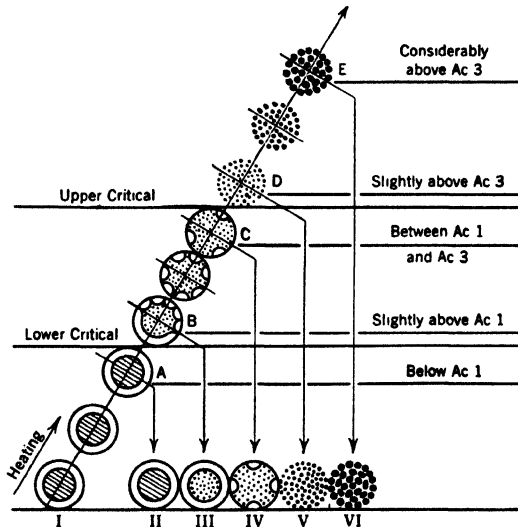


FIG. 38. Graphical illustration of changes in a 0.40% C steel on heating and quenching.

present. The most familiar is "18:8" steel containing 18% Cr 8% Ni. In austenitic steels of borderline composition, cooling below room temperature may produce martensite, and in such steels cold working may also produce it. With these exceptions, austenitic steels are not susceptible to heat treatment. Their consideration will be deferred to Vol. III.

**Heating** is a preparation for cooling. Austenite must be produced by heating before it can be changed into martensite by quenching.

Figure 38 illustrates the elimination of coarse grain size, network, free ferrite (in hypoeutectoid steels), and the necessity for complete diffusion and equalization of austenite.

**Changes on Heating.** In Fig. 38, column I represents an 0.40% C pearlitic steel (at the bottom of the column) and the structural

changes taking place in that steel as it is heated progressively through and beyond the critical ranges.

Thus, heating to a temperature *A*, under that of the lower critical range will produce no change\* in the original steel, which consists of pearlite (the inner cross-hatched circles) and ferrite (the white area), with the outer circumference representing the grain size. The quenching likewise will produce no change, as is illustrated by column II.

Heating to a temperature *B*, slightly above the lower critical range, will change the pearlite to austenite (represented by the dotted area), without affecting the free ferrite. Quenching, column III, will therefore produce a semihardened steel—since the solid solution is the “hard” constituent—with a refinement only of the “groundmass” (the original pearlite).

Heating to a temperature *C*, between the lower and upper critical ranges will effect a progressive absorption by the solid solution of the remaining free ferrite. Quenching, column IV, will therefore produce a “harder” steel than in case III, but nevertheless without complete refinement of the steel as a whole. This is indicated in the chart by the solid line of the circle, representing the initial grain size, and by the islands of free ferrite along the boundaries.

Heating to a temperature *D*, slightly above the upper critical range, if sufficiently prolonged to effect complete diffusion and equalization, will entirely transform the steel, giving it the smallest austenite grain size possible. Effective quenching, column V, will produce martensite and give the maximum hardness possible. The absence of the definite circumferential line indicates graphically the elimination of the initial grain size and the production of many fine grains within the area originally represented as a grain.

Heating to a temperature *E*, considerably above the upper critical range, will tend to increase the grain size; and quenching, column VI, will produce martensite, whose properties, even after tempering, will reflect the change in grain size.

#### PHYSICAL PROPERTIES DEPENDENT ON CONSTITUENTS

Starting with slowly cooled steel, the physical properties of the steel will depend on the relative proportions of the structural constituents, their relative arrangement being neglected for the time

\* Spheroidization can be produced below  $A_{c1}$ , but the time required is long.

being. Each of the components—ferrite, pearlite, and cementite—has certain physical characteristics.

**Ferrite.** Ferrite is soft, ductile, and relatively weak. Commercially pure ferrite has a tensile strength of approximately 40,000 to 50,000 psi with an elongation of about 40%. Such ferrite has a hardness of only about 90 Brinell. It is magnetic and has a fairly high electric conductivity. Its appearance under the microscope has been shown in the groundmass of the 0.07% C steel of Fig. 17, that is, as polyhedral crystals in the low-C steels.

Extremely pure ferrite is even weaker. The National Physical Laboratory of England<sup>9</sup> has reported its properties as:

Tensile strength	20,500–22,500 psi
Proportional limit	4,500– 5,000 psi
Elongation, per cent	35–50
Reduction of area, per cent	100

**Pearlite.** A common occurrence of pearlite in slowly cooled steels is in the stratified lamellar formation, composed of alternate plates of ferrite and cementite. Normal pearlite, such as shown in Figs. 4 and 19, has a tensile strength of approximately 125,000 to 150,000 psi, with an elongation of about 10%. Its hardness is 250 to 300 Brinell, depending on the fineness of its structure. The finer it is, the stronger.

**Cementite.** The properties of cementite are very little known, with the exception of its great hardness, about 650 Brinell, and its brittleness, which is extreme. Its tensile strength should be around 325,000 psi, but it is so brittle that no one has been able to prepare pure cementite in test bar form. Were it so prepared it would actually show very low tensile strength, because so brittle a material cannot be uniformly loaded in tension, but it should have high compressive strength. Its ordinary occurrence in slowly cooled steels (with C greater than 0.83%) is either as a network, as at the top of Fig. 14, or as spines and needles.

**Static Strength.** We may sum up these facts in their relation to the static\* strength of slowly cooled (annealed) steel as follows: Free ferrite has minimum tensile strength with high ductility; pearlite has maximum tensile strength with low ductility; free cementite confers added hardness and a tendency toward brittleness; the steel tends to break apart along the network or the spines of cementite;

\* "Static" tests are tests made under a single gradually applied load. Tests made under rapidly fluctuating loads or with a load rapidly applied are termed "dynamic."

hence in hypereutectoid steels with such structure there is a consequent lowering of the tensile strength. By increasing the amount of pearlite in the steel, the static strength is increased, but there is a corresponding decrease in the ductility. As an increase in the amount of pearlite necessarily means an increase in the amount of C, the effect of increased C will be the same.

Figure 12 showed the shift in mechanical properties of slowly cooled steel with change in C content. Still greater variations occur between the strength of quenched and untempered, and quenched and fully tempered heat-treated steels, as shown in Fig. 13, but a general inverse relation between strength and ductility, such as appears in Fig. 12, still holds. The stronger and harder the steel, the less its ability for permanent deformation.

**Distribution of Structural Components.** The vast difference in the properties of the annealed steels of Fig. 12, or of normalized steels, as compared with the quenched and tempered steels of Fig. 13 resides in the distribution and arrangement of ferrite and carbide, the building blocks of all constructional steels. The distribution is more important than the composition of the blocks, for, in the quenched and tempered steels, tempered to the same structure and hardness, the ferrite can be plain or alloyed, and the carbide can be plain  $\text{Fe}_3\text{C}$  or one in which considerable Mn or other carbide-forming element substitutes for part of the Fe in the  $\text{Fe}_3\text{C}$ , without notably affecting the properties. *In sizes that quench to all-martensite*, the equally tempered martensites of plain C and of alloy steels produce practically indistinguishable mechanical properties. Alloying is to allow producing all-martensite in larger sections than is possible in a plain C steel. On the other hand, in annealed, or normalized or fully spheroidized steels, where the ferrite paths are wider and less obstructed than in tempered martensite, strengthening of the ferrite by alloying is easily distinguished.

Beside the features of structure so far mentioned, other factors, some of which are revealed by microscopic examination, affect the suitability of steel for heat treatment. These various factors are discussed next.

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## CHAPTER 4

### SUITABILITY OF STEEL FOR HEAT TREATMENT

**Carbon Content of Steel for Quenching.** Commercial heat treatment by quenching and tempering is seldom attempted with steels below 0.15% C. Conversely, it is considered that freedom from any brittle martensitic layer in welding is secured if the C does not rise above 0.15%.

This figure is reduced in some highly alloyed steels in which the alloy lowers the eutectoid percentage of C, notably in Cr steels. As an extreme example Gillmor<sup>1</sup> shows martensitic hardening, even on air cooling of small sections, in a steel of 0.05 C, 12½ Cr, and 1% Mo.

The reason why the low-C steels are seldom quenched is evident from Fig. 7. Extremely high cooling rates are necessary to wipe out the ferrite-pearlite upper transformation and preserve austenite for transformation at a lower temperature, and, even when austenite is so preserved and transforms to an acicular structure, the transformation product may be bainite rather than martensite. In view of the differences in cooling rates between surface and center of even small water-quenched sections, it is obvious that the martensitic hardenability would extend only through an extremely shallow layer.

Discussion has raged on the question whether C-free Fe can be transformed to what can properly be termed martensite, and this is still a moot question. Epstein<sup>2</sup> remarks that martensite may contain from 0.20% C up. Hess, Doty, and Childs<sup>3</sup> report that ¼-in. plate of AISI 8715 steel, spot-welded, has a brittle martensitic structure in the rapidly cooled zone about the spot. Even, say, in a 0.10% C steel, if a very rapid quench of a very thin section did produce a true martensite, as Bain's<sup>4</sup> curve in Fig. 39 hints is possible, the hardness of such martensite is low, and, on tempering sufficiently to relieve quenching stresses, the strength would be low compared to that obtainable at, say, 0.20% C. Epstein<sup>2</sup> summarizes the work of Bühler and others demonstrating the presence of high internal quenching stresses in very low-C steels, even though no martensite was formed. Williams' curve<sup>5</sup> in Fig. 40 is dashed below 0.20% to

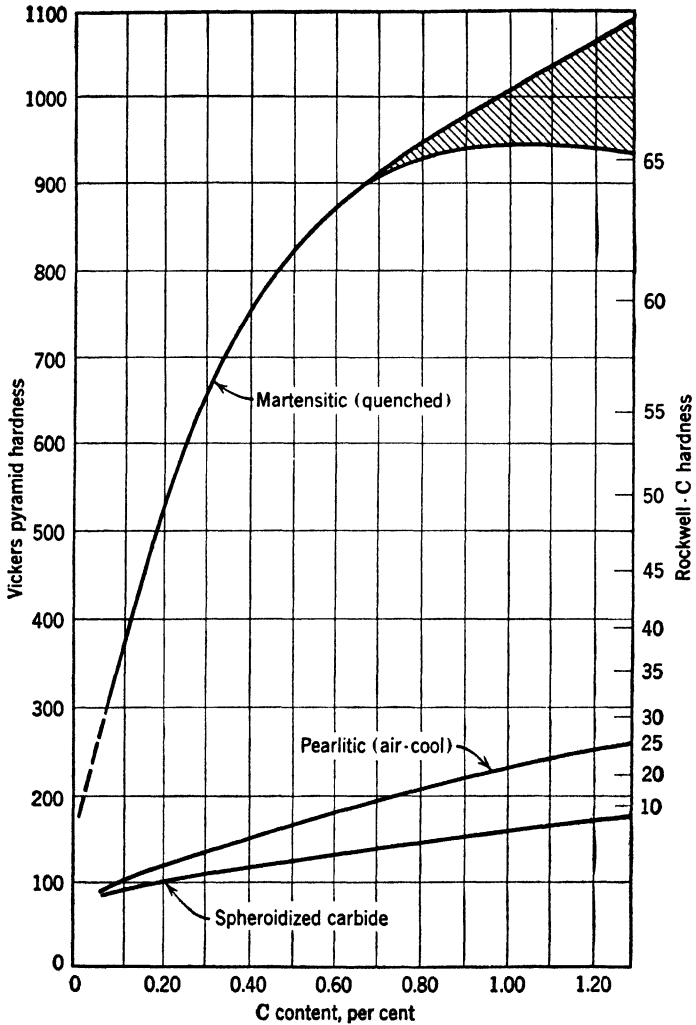


FIG. 39. The hardness of C steels in three structural conditions as a function of C content (approximate):

Top curve—Fully hardened martensite.

Middle curve—Lamellar structure (with proeutectoid constituents) as formed in normal cooling after rolling.

Bottom curve—Carbide coarsely spheroidized. Minimum hardness commercially obtainable.

indicate that the quenched structure may not be martensite. (Incidentally, Fig. 39 shows a possibility of higher hardness in quenched hypereutectoid steels than does Fig. 40. The upper boundary of the hatched portion of Fig. 39 is to indicate the conditions after retained austenite has been transformed to martensite by refrigeration.)

However, Fig. 39 indicates that a 0.10% C steel, with 100 Brinell as normalized, could, if rapidly cooled sufficiently, be brought to

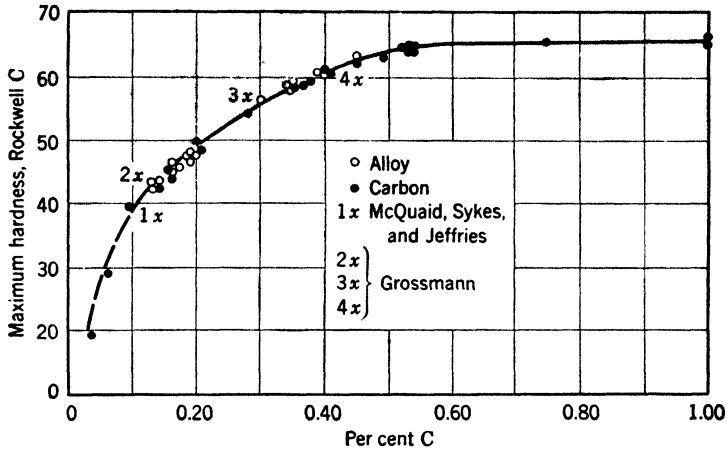


FIG. 40. Maximum hardness obtainable in steel is a function of its C content rather than of a moderate amount of alloying elements.

around 300 Brinell. But, as is shown in connection with an “S curve”<sup>6</sup> for a 0.12% C steel, forcing the transformation of a tiny piece to occur as low as 700° only produced 200 Brinell.

In a comprehensive study of S and P in rivet steel carried on by an ASTM committee,<sup>7</sup> a couple of rimmed steels, of normal S and P, containing 0.11 C, 0.40% Mn, and 0.13 C, 0.43% Mn, were used as base lines. These were normalized or water-quenched without tempering, in about 1 in. diameter and machined down to test specimens. They gave the results shown in Table 3.

TABLE 3

	Normalized	Quenched
Tensile, psi	55,000	62,000-70,000
Yield, psi	35,000-40,000	40,000-45,000
Elongation, per cent	40-45	30-32
R.A., per cent	72	72
Brinell	90-95	135-150

Smith<sup>8</sup> reported that a high-pressure water quench on a 0.505-in. tensile bar of 0.13 C, 0.51% Mn steel gave 100,000–120,000 psi tensile, 40% R.A. when quenched from 1700°, and that by pressure quenching commercial bolts and rivets could be produced from such steel to minimum specifications of 80,000 tensile and 15% elongation.

**Surface Behavior.** One might think that even a very thin harder surface layer produced by quenching a low-C steel, even in large sections, might be useful for wear resistance, but Rosenberg<sup>9</sup> found very poor resistance to abrasion by sand (by the little used Brinell abrasive-wear test) in a steel of 0.08 C, 0.48 Mn, 0.02% Si, water quenched from 1725° in  $3 \times \frac{1}{4}$  in., and apparently untempered. It had a Brinell of about 160. Later, he studied metal-to-metal wear in the Amsler wear-testing Machine, the test specimen being of 0.12 C, 0.44 Mn, 0.25% Si, water-quenched from 1700° in 2 in. diameter by 0.4 in. thick, tempered at 500°, and finish-ground. The surface varied from 21 to 37 Rockwell C, averaging 32. This would indicate irregular response to quenching, perhaps due to oxidation on heating, or to warping such that finish grinding cut through the surface layer to varying distances. Rosenberg noted signs of strain hardening in the microstructure, as though severe quenching stresses were set up and not relieved by the low-temperature tempering.

In spite of relatively high surface hardness the wear resistance, under the test conditions, was very inferior.

When a surface is to be made wear-resistant by surface hardening by flame or induction, a plain C steel of 0.40 to 0.60%, or, even when considerable alloying elements are present, usually over 0.30% C, is used. In plain C or low-alloy steels for case carburizing, which run from 0.10 to 0.20% C, the uncarburized core is generally considered not to be converted to martensite on quenching.

Even in steels with  $1\frac{3}{4}$  Ni,  $\frac{1}{4}$  Cr,  $\frac{1}{4}$ % Mo, at the 0.15% C level, the hardenability is low, as Rowland and coworkers<sup>10</sup> show.

Thus, in any, save rather highly alloyed steels, one can set 0.20% C as an approximate dividing line between the lower-C steels, not amenable to useful martensitic hardening on quenching, and those that respond to quenching. The treatment usually chosen to give the lower-C steels maximum useful strength is cold working, or normalizing to produce fine pearlite.

**Welding.** However, the response of the tiny "heat-affected" zone near a weld, to the very rapid cooling due to the large mass of cold metal in direct contact with it, is sufficient to bring in danger of cracking at that point, in a concealed location, through introduction

of cooling stresses or transformation at a relatively low temperature. The transformation is not necessarily to martensite or lower bainite; but, to whatever product the transformation is, it involves a change in volume at a temperature at which the steel is too stiff to accommodate the change. Even in steel with scarcely enough Mn to class it as an alloy steel, cracking is possible at a C level as low as 0.15 to 0.20%. The weldability of such steels is a major problem because they are on the border line between satisfactory lack of response to rapid cooling and a response that leads to unreliable welds. With suitable ordinary welding practice, the 0.15% C mild-alloy steels (described in Vol. III) and 0.20% C plain C steel are considered relatively fool-proof. By taking special precautions in welding, these limits can be shoved upwards.

#### KINDS OF STEEL, AS TO METHOD OF MANUFACTURE

Several different melting and "finishing" methods are used in making steel. Such methods may affect the way the steel has to be handled in heat treatment. The behavior of three low-C steels on annealing after cold working, reported by Wallace and Ricketts,<sup>11</sup> Fig. 133, p. 237, is an example.

**Rimmed or Killed Steel.** Practically all steel intended for heat treatment by quenching and tempering is "killed" by adding, shortly before pouring, sufficient Si, Al, Ti, Zr, or other "deoxidizers"<sup>11</sup> to remove the oxygen that is dissolved in the bath as Fe and Mn oxides and to put it into an insoluble form. Such steel lies "dead" in the mold and, as it freezes, contracts to form a "pipe." The molds normally used for killed steel are smaller at the bottom than at the top, and the top is surmounted by a "hot top," which keeps the top metal molten while the bottom is freezing, so that molten metal is fed into the central cavity formed as the metal freezes and takes up less space. The final cavity at the very top is cropped off. If the ingot has its big end down, there is likely to be a concealed pipe toward the bottom.

Rimming steel is poured when sufficiently high in O for a little further reaction of C with O or iron oxide to form the gas, carbon monoxide (CO), while the steel is freezing, thus producing bubbles or blowholes. The first metal to freeze on the bottom and sides of the mold is relatively pure iron and free from blowholes, both C and oxide being rejected to the liquid steel in the center. Thus the inside becomes a bit higher in C and O than when it was poured, and this

increase starts the reaction to CO. A part of the resulting gas is caught in the freezing metal, the aim being just to balance the volume contraction from liquid to solid steel, with the volume of gas caught as bubbles. These bubbles must be sufficiently deep-seated so that they will not break through but will remain clean and weld during the rolling of the ingot. Thus the interior of a piece of rimmed steel is of slightly different composition and of less dense structure than the outside. Since a relatively high O content is necessary for a properly controlled rimming action, steels very high in C or Mn or with much of any Si cannot be rimmed, since the presence of large amounts of these removes O from the molten steel.

Rimmed steel is prone to "age" (see Chapter 10), but it is claimed that aging can be prevented by a small addition of V, while the rimming action is preserved. Even relatively small additions of the very strong deoxidizers are, however, incompatible with rimming. Even a weak deoxidizer such as S is detrimental to rimming if present in, say, over 0.05%. Elements that have no deoxidizing action, Ni, Cu, and Mo, are compatible with rimming, but alloy rimming steels are rarely made, except for Cu steels aimed to show resistance to atmospheric corrosion. Because of the low-C level, rimmed steels are seldom heat-treated by quenching and tempering, though at the highest levels of C and Mn allowable for rimming, there would be some response to such treatment.

Rimmed steel is less expensive to produce than killed steel because of the larger discard in cropping the pipe from the latter. Rimmed steel finds much use for stampings and the like where its superior surface is an asset. Annealing of cold-worked and normalizing of hot-rolled material are the heat treatments of major importance for rimmed steels. Ship plate has usually been made from rimmed steel. Welding, which inescapably involves heat treatment of a sort, is done on rimmed steel, perhaps more often than on killed. Steel castings, of course, are always killed, since they should be sound, and so rimmed steel is met only in wrought products, usually being limited to plates, sheets, and wire.

"Semikilled" steel is made at a slightly lower O level than rimmed steel; whereas rimmed steel is practically Si free and has had only a trace of Al added, semikilled carries a larger "trace" of Al and around 0.02–0.10% Si. From the heat-treatment point of view, it has much the same limitations as has rimmed steel.

**Acid and Basic Steels.** Steels may be made in furnaces whose bottoms and side walls are lined with silica (acid) or magnesite

(basic). Basic slags, high in lime, cannot be used with silica bottoms, as they eat the silica away very rapidly, so the purification that is accomplished by basic slags cannot be carried out within the acid furnace; hence the scrap for acid melting has to be relatively free from those impurities removable only by basic slags. Nonmetallic slag inclusions in steel are naturally of varying composition and behavior, according to the type of slag used, so that, even though the chemical analyses of the steels themselves are alike, an acid and a basic steel may not be alike in all respects. However, what differences there are are seldom important from the point of view of heat treatment. Most of the inclusions in steel are not trapped slag, but are products of the deoxidation reactions, which are universally present, along with manganese sulphide.

Open-hearth and electric furnaces may be either acid or basic. In the wrought-steel industry, basic linings are by far the more common. In the steel-casting industry, acid linings are more common. The Bessemer converter can be either acid or basic, but in American practice it is universally lined acid.

**Bessemer, Open-Hearth, Electric, and Duplex Steel.** The acid Bessemer converter takes high-C metal previously melted in the blast furnace or the cupola and makes steel by blowing air through it or on its surface. The molten metal supplied is usually rather high in Si, and the heat is largely supplied by combustion of this Si and of the C in the melt. No outside heat is used. The process is rapid; a large heat is made in about a quarter of an hour. Heats of 25 tons are common, but recently installed converters ordinarily contain 40 tons. The metal comes out low in C and with O content corresponding to it, the C:O ratio being the same as in open-hearth metal of like C content. Phosphorus is not eliminated in the acid converter, without special treatment, so that the P is ordinarily around 0.10% instead of around 0.02–0.04%, as in the open hearth. Phosphorus is a strengthening element, but, with P at 0.10%, C cannot safely rise much above 0.15%; else brittleness is met. This level of C is too low for heat treatment by quenching and tempering. Such steel can be case-carburized and then becomes heat-treatable. (The high-C high-P case is brittle, but so is the high-C low-P case.) The P content can be reduced to 0.05% or below by treatment outside the converter, with a basic slag, as the steel flows into a basic lined ladle, without injuring the rimming characteristics.

Bessemer steel made in the bottom-blown type of converter also differs from open-hearth in that it contains around 0.01 to 0.02% N,



which increases the propensity toward "aging." Killing the steel with elements that combine with N and put it in an inactive form, and that similarly bind the O greatly reduces the propensity toward aging. A killed converter steel may be raised to any desired C level and can be alloyed to whatever degree can be accomplished in the ladle. The properties then differ slightly from those of open-hearth, according to whether dephosphorization has been practiced, and to what degree, and according to the content of nitrides.

Killed Bessemer steel, especially the dephosphorized type, brought to a suitable C level, would be heat-treatable and will probably be so used in the future, though most present uses for this type do not involve quenching and tempering. Killed converter steel, made from low-P scrap and in the side-blown converter, which blows the air onto the surface of the melt instead of through it, thereby avoiding so much pickup of N, is quite as usable for castings, and quite as responsive to heat treatment as open-hearth or electric steel of like composition.

The open-hearth furnace is heated by flames playing over the bath, but kept from contact with the melt by the layer of slag. Nitrogen pickup is small, and in the basic furnace, both S and P can be largely removed by suitable slags. Most open-hearth furnaces for wrought steel have large capacity, 100 to 200 tons, and the heats take 6 to 12 hr. During the heat the same sort of oxidation process goes on as in the converter, only much more slowly. It is possible to stop the process at a higher C level than in the converter, but, in general, oxidation is carried on nearly to the same degree as in the converter.

Oxidizable alloying elements, such as Cr, are at least partly removed by oxidation and transference of the oxide to the slag. This alters the behavior of the slag and complicates control. Nonoxidizable metals, Cu, Ni, and Mo, are not removed. (This is true even in the converter.) Alloy steels high in Cr are therefore seldom made in the open hearth, though many of the low-Cr N.E. "triple-alloy" steels can be satisfactorily made in it. The bulk of the plain C steel for heat treatment is made in the open hearth.

Electric-arc furnaces can be operated at higher temperatures than the open-hearth, and the temperature and the slag composition are under rather more precise control. They can be operated under oxidizing or nonoxidizing conditions, depending on the type of slag employed. By skimming off the slag at the proper temperature and replacing it with a different type, purification by removal of some of the unwanted impurities and retention of the bulk<sup>4</sup> of even the

oxidizable alloying metals can be accomplished. High-alloy steels containing expensive oxidizable alloying elements are therefore chiefly made in basic electric furnaces. Some such furnaces hold up to 75 or 100 tons, in the wrought-steel industry. A large proportion of the steel castings are made in the acid electric furnace, the capacity usually being 3 to 10 tons.

The induction electric furnace may, with some difficulty, be operated with refining slags. It is more commonly used without a slag, as a remelting furnace. There is then very little loss, even of oxidizable elements. It is most used in 1- to 3-ton sizes.

The high temperature attainable in either type of electric furnace allows easier production of clean steel, though dirty steel can be made in them and electric melting does not insure that the steel be clean. However, it is considered decidedly easier to produce "air-craft-quality" steel from an electric furnace.

**Duplex Steel.** Of course it is possible to melt by one method and finish by another. For example, Bessemer steel may be transferred to the open hearth, or to the electric. Or open-hearth steel may be finished in the electric. Such mixed processes are called duplexing. Much duplex steel, made in the period now well past, had a blend of, or more or less averaged, the characteristics of steels made by the separate melting processes. However, the present tendency in duplexing is to consider converter metal as a hot metal charge and to supply sufficient C in one form or another, so that the necessary active boil in the open hearth is long enough continued to carry out most of the N brought in by the converter metal, and to produce steel of the usual open-hearth degree of cleanliness. For some uses, the properties and behavior of such duplex steel are still distinguishable from those of regular open-hearth metal, but, for most purposes, are quite adequate.

Machinability, fatigue resistance, directional properties, and other features related to type and distribution of inclusions are or may be related to the melting process, and these features may be important in the service of the heat-treated part, and hence may affect the selection of the steel to be heat-treated, according to the method by which it has been melted.

But from the heat treater's point of view, little attention need be paid to the melting process by which the steel has been made. The technique used in finishing the steel, in respect to grain size and hardenability control, produces more marked differences in response to heat treatment than does the melting process.

## OBJECTS OF TESTING

In order that the producer and consumer may speak a common language in discussing the quality of steel, it is necessary that "quality" be described, not merely in vague or qualitative phrases but, wherever possible, in definite figures. This leads to mechanical and physical testing, first for the selection of suitable material, second for its duplication. The effects of heat treatment on steel of given chemical composition as well as the proper correlation of composition and treatment are appraised by such testing.

The heat treater is interested in tests that will enable him to tell the steel maker what he wants in the steels he uses and in those tests which, in turn, enable him to satisfy the user of the heat-treated part that it has the properties which the user demands. In both cases definite specifications are more helpful than vague phrases, *provided always, that the specifications really define the properties actually desired.*

## TESTS ON MATERIAL FOR HEAT TREATMENT

Chemical analysis used to be the primary specification in the selection of steels for heat treating. But the question, "Is this lot of steel fit to be heat-treated for the purpose in hand?" cannot be answered by a mere chemical analysis. The C level is of importance, and knowledge of it is helpful, since it is a main determining factor in the selection of temperatures for heat treatment. But there are more direct methods of evaluating the attributes that used to be roughly inferred from the analysis.

Steel fit for heat treatment needs, besides the proper C level, to have:

1. Its nonmetallic inclusions of suitable amount, type, and distribution.
2. A suitable degree of initial homogeneity.
3. Known propensity to or away from grain growth at heat-treating temperature and, if it is to be cooled at a controlled rate and then tempered;
4. The desired degree of hardenability.
5. The desired rates of transformation at different temperatures.
6. A known response to tempering after hardening.

It may also be necessary to determine:

7. The "critical temperature," above which the steel must be heated for complete hardening, or, more specifically, the time and temperature for homogenization on heating should be known.

8. The propensity toward decarburization is also a factor.

With information on these factors, the heat treater can eliminate unsuitable steels and adapt his heat-treatment practice so as to produce the desired structure and hence the desired properties. All these factors can be specified in ordering the steel, or they can be tested for on any shipment.

The first two factors are discussed here, the others being left for later chapters.

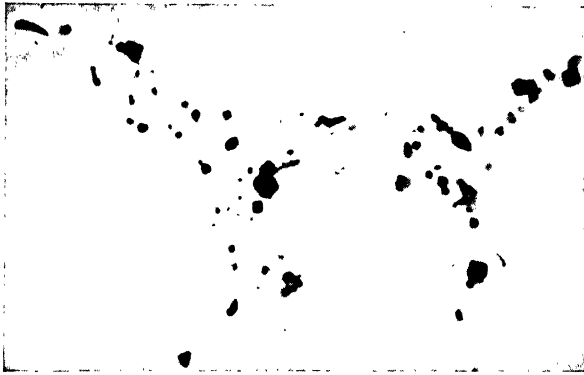


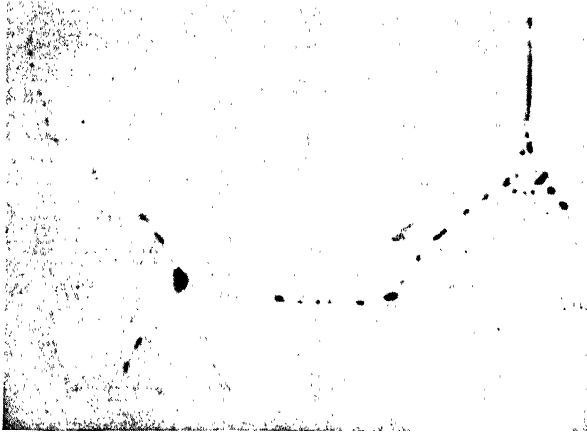
Fig. 41. This "bird dog on point" is composed of colonies of oxide and sulphide inclusions at the grain boundaries in cast steel. Unetched. (Battelle)

**Inclusions.** Metallographic technique is used, not only to reveal the structure of the steel itself, but also to show the presence, distribution, and type of inclusions.

Inclusions are not always bad. They may definitely be wanted. Manganese sulphide inclusions are present in all steel, since S is never completely eliminated. In cast steel, this brittle nonmetallic substance can occur, either as films or as globules between the metallic grains (Figs. 41-43).

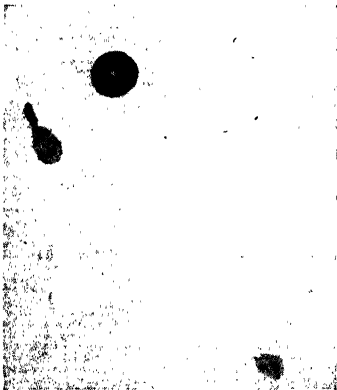
The properties of the steel after heat treatment are naturally affected by the mode of distribution. The film distribution is to be avoided and is avoidable.<sup>12</sup> When steel is hot-rolled or forged, the originally globular manganese sulphide inclusions, which are plastic at hot-working temperatures and located at or near grain boundaries, are spread out into plates interspersed through the metal (Fig. 44). These inclusions, brittle when cold, are not very abrasive to cutting

tools, but they allow the tool to cut short chips without tearing the metal deeply. They increase machinability and promote smooth finish, but do so most effectively when they are relatively small and uniformly distributed, rather than very large and scattered irregu-



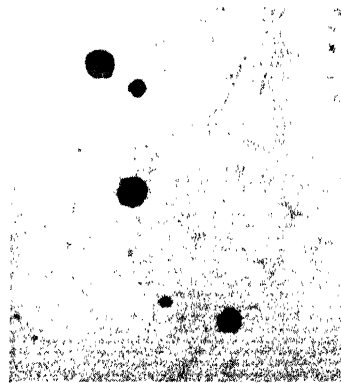
× 500.

FIG. 42. Film- or stringer-type single-phase sulphide inclusions found in cast steels receiving insufficient additions of a strong deoxidizer. Unetched. (Boulger, Battelle)



(a) × 500.

FIG. 43a. Impure silicate and globular duplexed sulphide inclusions in a silicon-killed grade-B cast steel. Unetched. (Boulger, Battelle)



(b) × 1000.

FIG. 43b. Relatively pure globular manganese sulphide inclusions found in a silicon-killed grade-B cast steel. Unetched. (Boulger, Battelle)

larly. In certain heat-treated gears, the life is vastly increased when the machined finish is smooth, and to secure this smoothness the S content is held at about 0.05% in order to supply the needed amount of manganese sulphide in a steel that would otherwise ordinarily be specified to have 0.03% maximum. The distribution of the inclusions is also carefully watched. The toughness, especially in the transverse direction, is somewhat decreased by the larger



Fig. 44a. Longitudinal view of single-phase sulphide inclusion in hot-rolled steel. Unetched. (Boulger, Battelle)



Fig. 44b. Longitudinal view of duplex silicate-sulphide inclusion in hot-rolled steel. Unetched. (Boulger, Battelle)

amount of inclusions, but the gear cannot be allowed to take a permanent set, for that would spoil it, and so toughness is not a primary factor. In this case the steel, in spite of its inclusion content, has adequate toughness for "accident insurance." But very high S is likely to lead to very poor toughness in the transverse direction.

**Hard or Soft Steel.** On the other hand, in steel heat-treated to a high level of hardness, and in a part subjected to higher repeated stresses, large inclusions at or close to the surface may be fatal. Figure 45 shows the type and size of inclusions in a heat of steel purchased for manufacture of leaf springs. Springs made from this failed promptly in fatigue at around two thirds of the stress that a normally clean spring steel will endure indefinitely. Similarly, in steel for ball or roller bearings, inclusions are highly detrimental, since such bearings are very hard.

On the other hand, in very soft rivet steel large amounts of inclusions have no perceptible effect on endurance. The harder the steel is, the more likely it is that inclusions will be detrimental in service.

**Varieties of Inclusions.** The manganese sulphide inclusions and those largely consisting of that sulphide in wrought steel are strung out in the direction of working (Figs. 44 and 47), and, if the stress applied to the finished part in service is also in this direction, such inclusions have little effect. If the steel must deform plastically in

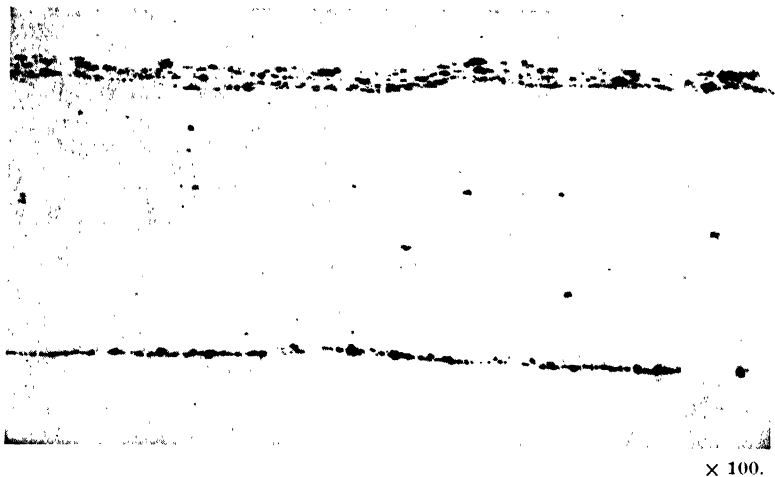


Fig. 45. Longitudinal stringer-type inclusions, damaging to the fatigue strength of a Mn-Mo spring steel. Unetched. (Fick, Battelle)

the cold and is stressed *across* the inclusion direction, its ability to deform is reduced, and directional properties ensue (see Chapter 5).

Inclusions that are not plastic at hot-rolling temperatures, such as aluminum oxide (Fig. 48), titanium nitride and zirconium nitride (Fig. 49) are either naturally sharp cornered, or break up into sharp-cornered particles. The fragments may be strung out into relatively adjacent locations, more or less in line (Fig. 46). The individual particles are "stress raisers," and the series of particles acts as stepping stones for the progress of a crack. These sharp hard inclusions act to increase the stress concentration in any direction and are especially deleterious when strung out. Such inclusions are not desired in "aircraft-quality" steel.

Not only are they detrimental in service, but they may help produce that bugbear of the steel treater, quenching cracks. Fortunately, large aluminum oxide inclusions are not necessarily present

in Al-treated steel; the aluminum oxide may be engulfed in the manganese sulphide particles, any free aluminum oxide being present only in nearly or completely submicroscopic size.



× 500.

FIG. 46. Same as Fig. 45 but at higher magnification. Note the sharp corners on the inclusion in the upper center. Unetched. (Fick, Battelle)



× 250.

FIG. 47. Longitudinal view of duplex silicate-sulphide inclusions in high-S hot-rolled steel (0.40 C, 1.5 Mn, 0.25 S, 0.20% Si). The dark portions, which are largely silicate, are the more plastic. Unetched. (Boulger, Battelle)

Silicate inclusions are not so hard, sharp, and harmful as aluminum oxide, and may also be engulfed (Fig. 46) in manganese sulphide. They are usually plastic, but sometimes are brittle, as in Figs. 49 and 50. A good account of the nature and genesis of inclusions is



given in the AIME volume on open-hearth steel <sup>18</sup> and another by Rait and Pinder <sup>28</sup>

**Examination of Inclusions.** It is possible to extract the inclusions from a large sample of steel by chemical methods and thus determine

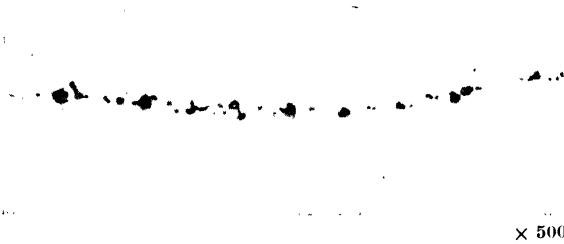


FIG. 48a. Streak of impure alumina inclusions, longitudinal view, in  $1\frac{1}{8}$ -in. plates of 4330 steel. Unetched. (Boulger, Battelle)

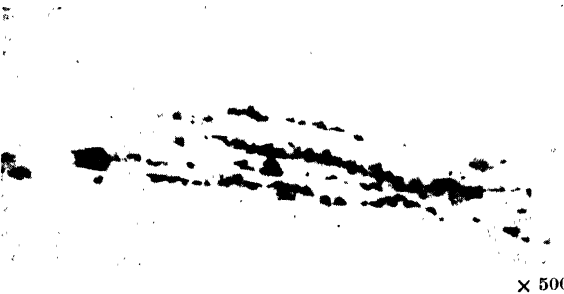


FIG. 48b. Another streak of impure alumina inclusions, longitudinal view, in  $1\frac{1}{8}$ -in. plates of 4330 steel. Unetched. (Boulger, Battelle)

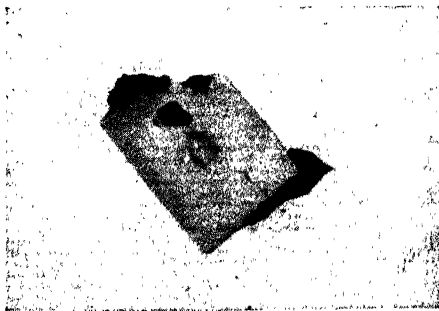


FIG. 49. Crystal of zirconium nitride found in a wrought steel containing 0.09% Zr. The crystalline outline shows that the nitride was not malleable at rolling temperatures. (Boulger, Battelle)

the amount. The separated inclusions can in turn be microscopically examined and subjected to chemical analysis, if we are interested

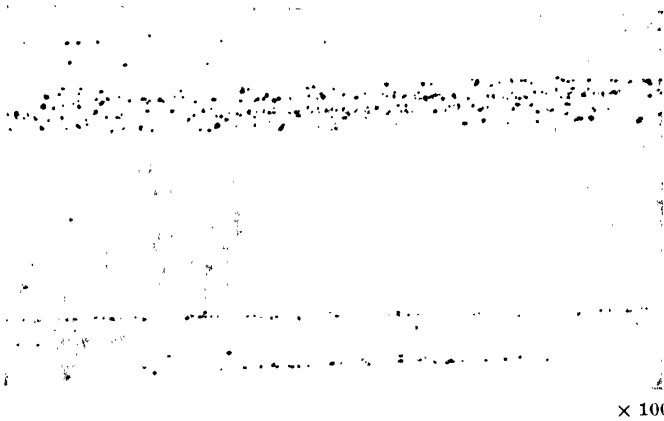


FIG. 50a. Refractory impure silicate inclusions found in streaks in  $\frac{1}{2}$ -in. hot-rolled plate of commercial steel. (Rare; ordinarily silicates are plastic at steel rolling temperatures.) Unetched.

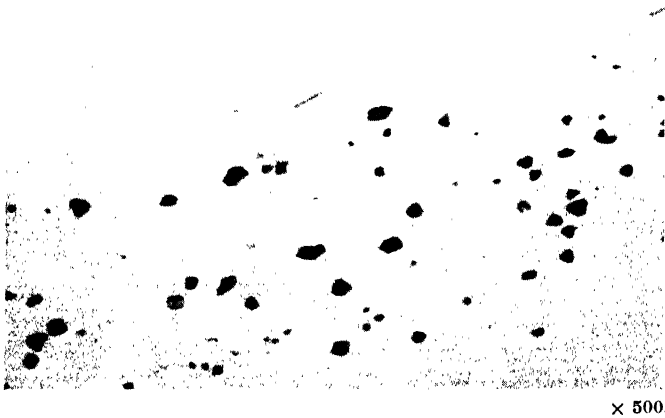


FIG. 50b. Same as Fig. 50a, at higher magnification. Unetched. (Boulger, Battelle)

in their origin, but such procedure is too lengthy for an acceptance test.

Microscopic examination of polished unetched sections appraises the steel for cleanliness, if many sections are examined to sample the steel adequately, but adequate sampling is tedious. However, this method is much used, and various users have set up their own stand-

ards as to amount, type, size and distribution of inclusions in "rating charts." The ATSM chart (E45-42T) is the most familiar.

Figure 51 shows the end members of the ASTM series, illustrating sulphide inclusions. These represent steels, respectively, very low and very high in S, as they appear when a polished unetched section is viewed at  $100\times$  under the microscope. A quicker way to get adequate sampling is to smooth-turn a specimen and "magnaflux" it. Large inclusions or strung-out clumps at or close to the surface are thus revealed. Conical or stepped specimens, or successively turning off deeper cuts and then magnafluxing, will indicate the content of large inclusions at varying depths. Inclusions revealed by magnafluxing are ordinarily of such size as to affect the properties, especially the transverse properties. Much experience and judgment are required in appraising the proper cleanliness rating for a given part, but, when the decision has been sanely made, steel too dirty for the service in view can be discarded without wasting money on its heat treatment or machining.

**Homogeneity.** Medium-C steels, those most used for heat treatment, develop a duplex structure on slow cooling or normalizing, with the soft nearly pure Fe constituent enveloping pearlite, the harder conglomerate of Fe, and its carbide. When large units cool slowly this structure may occur on a gross scale, and, when the metal is elongated in rolling or forging, the pure iron sheaths are squeezed into bands. This banding is the more marked, the higher the P, and the slower the cooling. It seems also to be more marked in steels high in Mn.

**Banding.** This condition can be traced back to the heterogeneous alloy distribution resulting from dendritic segregation, which occurs during the freezing of the ingot. While C diffuses quite rapidly, such alloys as Mn, Ni, and Cr diffuse very slowly at the temperatures normally used during rolling, so that the alloy segregation persists throughout processing. While this segregation is on a microscale, it results in areas both lower and higher in alloy content than that of the average analysis. During slow cooling, such as normalizing and especially annealing, the carbon diffuses rapidly to the high-alloy areas, which remain austenitic for the longest period. This finally results in the formation of pearlite in the high-alloy bands.

A banded structure may increase the tendency for differences in directional properties, and the heat treater is often charged with the task of remedying a banded structure by diffusion at high temperature (Fig. 52).

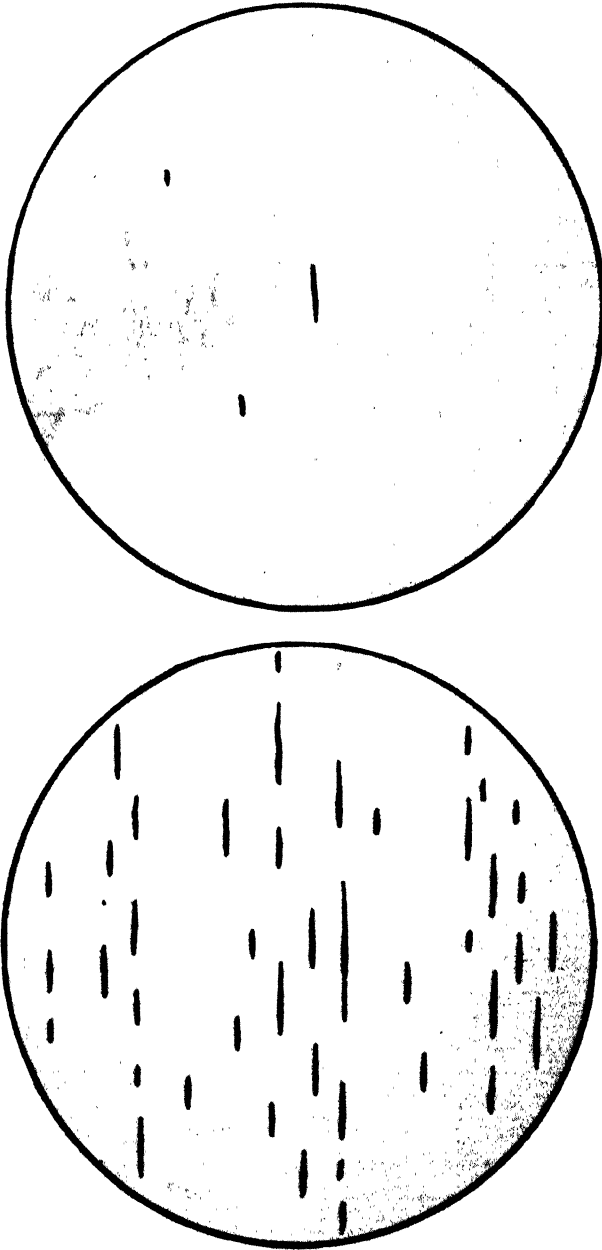


FIG. 51. Samples of the ASTM inclusion rating charts, heavy series, thickness up to approximately 6 microns (0.00025 in.).

Under the microscope, banded steels usually look worse than they would be judged to be by their behavior in mechanical tests, but there are sometimes legitimate reasons for remedying banding. This



Fig. 52a. Banded structure, after furnace cooling from 1600°. Note sulphide inclusions, elongated in the rolling direction, above arrow.

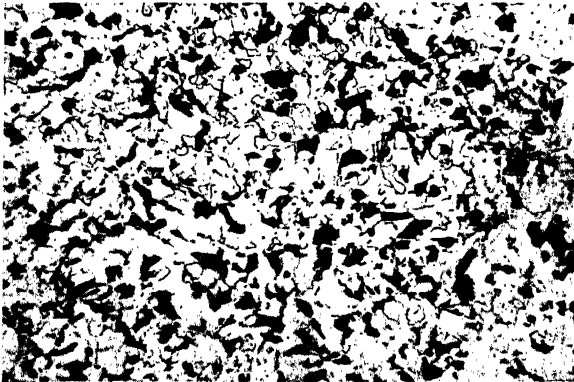
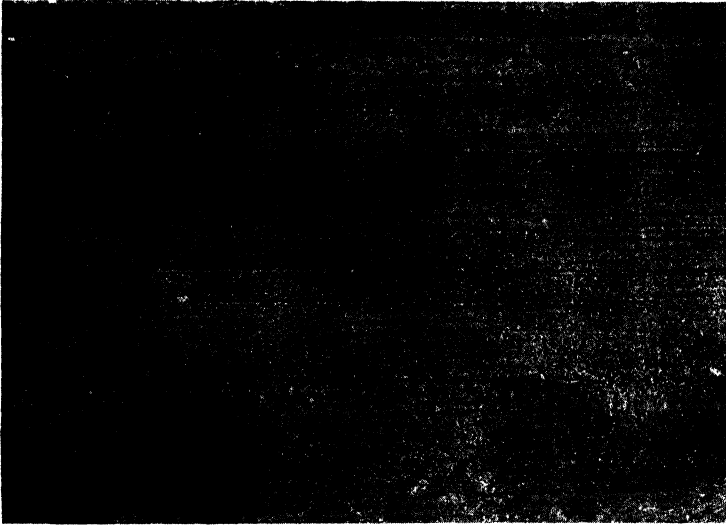


Fig. 52b. Same steel, homogenized 4 hr at 2350°, then furnace-cooled from 1600°. 0.17 C, 1.19% Mn steel. (Banta, Battelle)

task is sometimes difficult, and it may be wise to discard extremely banded stock rather than attempt to heat-treat it. The effect of banding on hardenability is discussed on pp. 275 and 289.

There is some segregation of C and of other elements in an ingot, and if the ingot has not been sufficiently cropped to remove the most segregated portion this variation may affect the behavior in

heat treatment. Alloying elements added in the mold or in the ladle at the last minute may not dissolve and diffuse uniformly. In extreme cases, such segregation may show up in etching a micro-section. Thus metallographic examination under the microscope may reveal features that make one lot of steel respond differently to heat treatment from another, or that lead to nonuniform response from piece to piece in a given lot.



× 1.

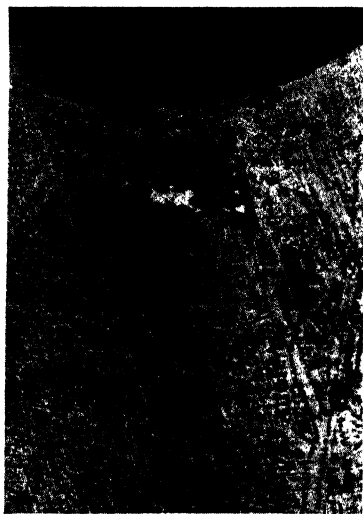
FIG. 53. Macroetched surface of hot-rolled bar showing numerous seams which were not apparent in an unetched specimen. Etchant—50% HCl at 170°F. (Boulger, Battelle)

**Deep Etching.** Ingot features that affect quality are the presence of concealed pipe or spongy centers in improperly fed ingots, a fault preferably avoided by the use of big-end-up hot-topped ingots. Unless the ingot has been cast at the proper temperature and properly heated for rolling or forging, and has had a sufficient degree of hot working, traces of the coarse dendritic cast structure may persist in the finished stock. Deep etching is helpful in detecting such conditions. It also reveals fiber direction in a forging, or in rolled stock, thus giving a hint as to directional properties.

Deep etching consists in attacking a specimen, usually a sawed-off cross section, by acid, usually about half commercial hydrochloric acid (perhaps with a small amount of sulphuric acid) and

half water, at about 160° for 15 min to 1 hr, depending on the type of steel.

Seams, pipes, cracks, spongy centers, large slag inclusions, and the like are enlarged and shown up in accentuated fashion by the acid attack. If the dendritic structure of the original ingot has not been broken down in the working of the steel, that structure may



× 4.

FIG. 54. Macroetched section showing shrinkage and nonmetallic inclusion near surface of a casting. Etchant—50% HCl at 170°F. (Boulger, Battelle)

be revealed, though whether such a structure is undesirable is a question only to be answered in respect to a specific steel and a specific service. Keshian<sup>14</sup> reports as good service from dies with a marked dendritic structure as from those free from it. Thum<sup>15</sup> notes that the dendritic structure in cast guns produces no harmful effect, whereas banding is injurious. Captured German wartime forgings<sup>16</sup> often showed marked dendritic structure, being made from blanks cast nearly to finished size and not given sufficient forging reduction to obliterate it.

Gross imperfections that exist but cannot be seen without etching are shown up, and, when these would be detrimental in service, the steel may be discarded, and the work may be avoided that would

otherwise have been put on it, only to have it fail to pass final inspection (or still worse, to slip through such inspection and come back, failed in service). The deep-etch test so exaggerates and magnifies inhomogeneities that the unskilled may be tempted to reject normal useful material on the basis of its looks after deep etching; so the test must be employed with discretion tempered by experience. Suggested practice for the test<sup>17</sup> and extended discussion of its application,<sup>14, 17</sup> are available in the literature. Figures 53–56 show deep-etched specimens.

**Decarburization.** Medium- and high-C steels, such as forging steels, spring steels, tool and die steels, are quite readily decarburized in heating for rolling or forging or in heating for normalizing or

quenching. Some of the C of the surface is oxidized away, as shown in Fig. 57. The soft decarburized "bark" is less hardenable than the body, and this soft skin must often be removed for the heat-



FIG. 55. Macroetched section showing slight microshrinkage along the center of a casting. Etchant—50% HCl at 170°F. (Boulger, Battelle)

treated part to function in the desired fashion. If the piece is to be heat-treated oversize and finish-machined or ground deeply enough to remove the decarburized surface with certainty, the presence of "decarb" is not serious. But, if it is not removed, its presence can be very serious. Thus, if a part made from drill rod is to be, in the finished piece, very near to its initial diameter, or, if a coil or flat spring is made from rolled or drawn stock without the



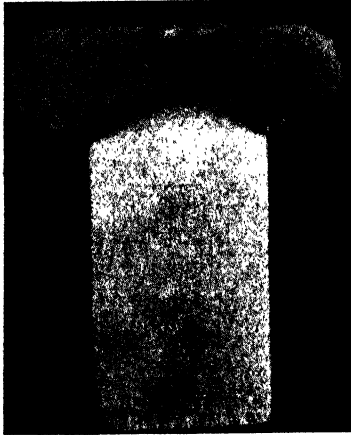
surface being removed, the decarburization which that stock may have, as it comes from the steel maker, may render it unfit for its purpose, or, if only a little is to be removed, a slight added decarburization from the heating for

quenching may be the straw that breaks the camel's back.

In tools and dies even the slightest decarburization is to be avoided, and it may be necessary to use salt-bath heating instead of furnace heating or to use a controlled-atmosphere furnace, in order to avoid it.

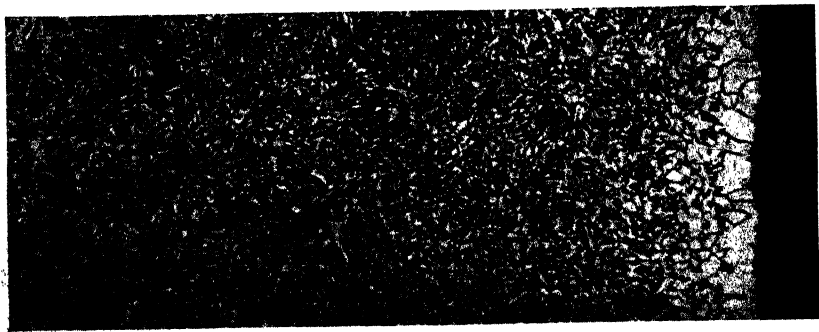
Mild cases of decarburization may sometimes be repaired, at least in part, by heating in a carburizing "controlled atmosphere" or in a cyaniding bath, and thus the heat treater can sometimes, to some extent, make up for the deficiency of the steel as received (see Vol. II).

Appraisal of the seriousness of decarburization can only be made with full knowledge of the type of service the finished part must endure and what machine work will be done in making the part. But



× 2½.

FIG. 56. Section through cold upset bolt. 0.30 C, 3.5% Ni. Deep-etched. (Pulsifer)



× 100.

FIG. 57. Decarburized surface on 0.60% C silicomanganese spring steel, produced during hot rolling and subsequent heat treatment. (Fick, Battelle)

much poor performance has been blamed on the heat treater that was his fault only insofar as he failed to appraise the incoming decarburized stock as unsuitable.

**Grain-Coarsening Characteristics.** The mechanical properties conferred by variations in grain size and its concomitants are discussed in Chapter 15. To anticipate, toughness is favored by small grain whereas behavior at very high temperature is usually better with coarse grain; machinability for some types of machining is so much better with coarse grain that the heat treater is often called on to produce coarse grain as a first step for machinability, and then again to treat the machined part to produce fine grain for mechanical properties.

The grain size in this connection refers to that grain size existing in the steel at the elevated temperatures used for normalizing or quenching.

There are two main types of steel in reference to grain size. If no special addition is made, as was the normal case 20 years ago, heating only a little above the minimum temperature for hardening caused rapid grain growth. Such steels are easily coarsenable. In those days a V addition was the only known method of producing fine grain. Many modern steels are made fine grained, that is, resistant to coarsening at quite high temperatures, by judicious addition of a small amount of Al, Ti or Zr (most commonly Al) or some combination of these. With such steels, instead of having to be very careful not to exceed materially the lowest possible hardening or annealing temperature, there is considerable leeway for working on the side of higher temperature (and considerable virtue in so doing).

But even the steels resistant to coarsening can be coarsened by going to a very high temperature, so that it is still possible to coarsen for machinability, or for better higher-temperature behavior.

Specifications for steels to be heat-treated now very commonly include a grain-size restriction by which a definite grain size after heating for quenching is called for.

Historically, grain size first came to be of interest in steels for carburizing, and the test method was to carburize for 8 hr at 1700°F, slow-cool, section, polish, etch, and examine. The carburization outlined the grains so they were readily visible and measurable. This was the "McQuaid-Ehn test," once much used and still employed.

The test did not always give valid results on steels to be heated at a lower temperature than 1700° or for less than 8 hr. Many steels that are fine grained with the usual time and temperature for quenching coarsen under the conditions of the McQuaid-Ehn test; thus the results are not interpretable and tend toward false conclusions. The test did pick out the more difficultly coarsenable steels,

for one that remained fine grained under its conditions would be fine grained under normal heat-treatment conditions.

Obviously the heat treater needs full information on the results that will spring from the quenching temperature, so instead of a single test, a series of specimens are heated without carburizing for a reasonable time to a range of temperatures, a specimen is quenched from each temperature, and the grain size of each is determined. Thus the coarsening tendency is revealed.

**Grain-Size Rating.** Different etching techniques on polished sections are required to reveal the grain size in different steels.<sup>18, 19</sup>

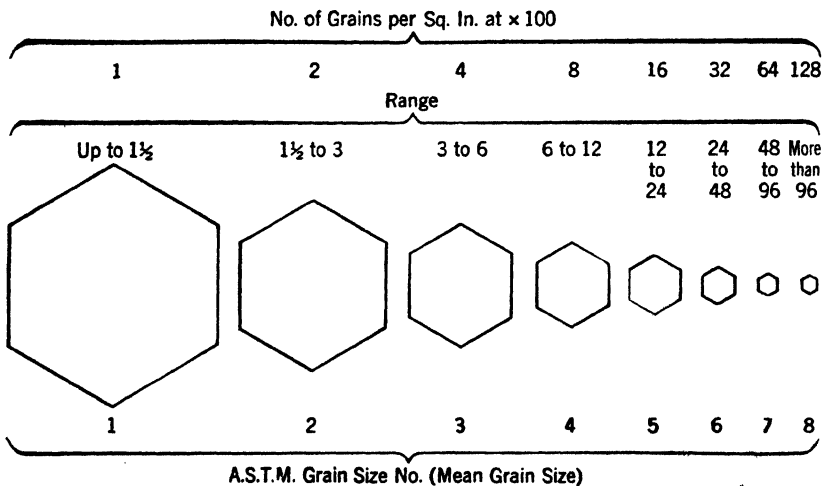
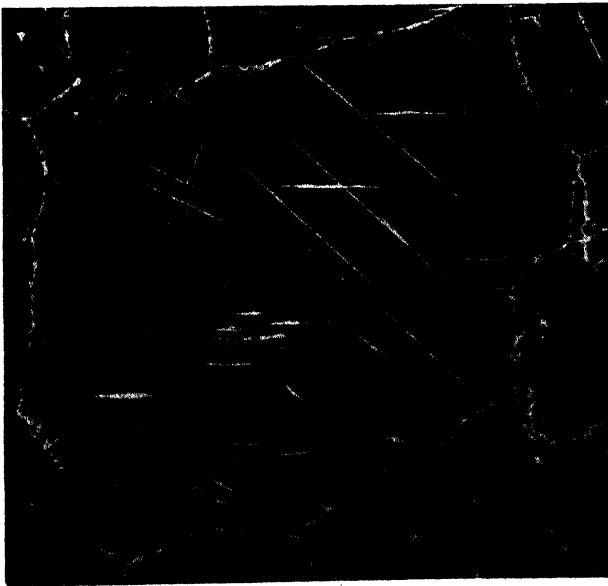
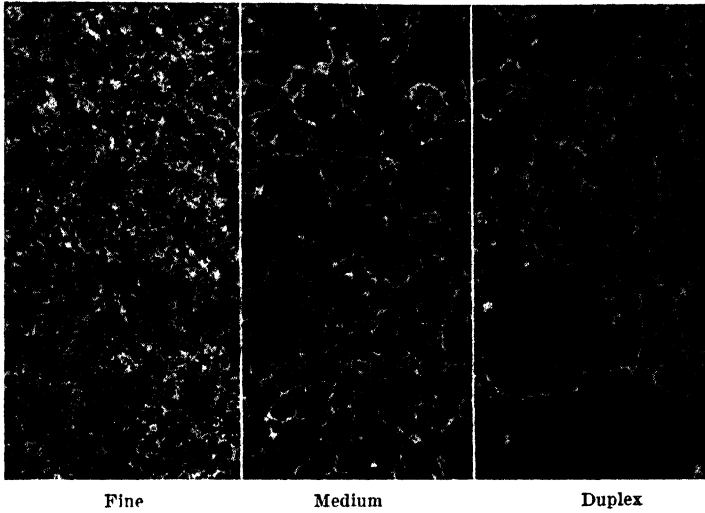


FIG. 58. Idealized grain size of steel as viewed at  $\times 100$ .

Once the grain size is revealed, it can be measured under the microscope and expressed in terms of area of grain, or more simply, by comparison with standard micrographs. The American Society for Testing Materials has put out a comparison chart, and grain size is almost universally expressed by ASTM numbers, Fig. 58. Figure 59 shows a range from very small to large grain size. In tool steels the grain size is often determined by a fracture test (p. 371).

**"Normal" and "Abnormal" Steel.** Still another differentiation between steels of nominally the same compositions relates to their degree of "normality." Historically, this distinction arose in relation to observations made on low- and medium-C steels that were case-carburized, especially when case carburizing was applied as a means of evaluating grain size. In the steels then available, it was common that the slow-cooled carburized case of a fine-grained steel



Very coarse. All  $\times 100$ .

(Lauderdale, Battelle)

FIG. 59. Hypereutectoid zones of different steels, all gas-carburized 8 hr at  $1600^{\circ}$  and slowly cooled. The grains are outlined by cementite.

would be found to have its pearlite grains appear broken up, and in coalesced masses, surrounded by ferrite, whereas in coarse-grained steels the pearlite grains looked as they would in a customary high-C steel, with cementite appearing in thin lines at the pearlite grain boundaries, with no appearance of free bands of ferrite. This was considered the usual or normal structure, and so it was called "normal," and the opposite structure, "abnormal." Figure 60 shows such structures.

Fine grain in the carburizing test and abnormality were, for a time, thought to go hand in hand, but it soon turned out that fine-grained normal steels could also be produced. The usual cause of "abnormality" is generally thought to be very complete deoxidation of the melt. Elimination of the last traces of iron and manganese oxides, as by a relatively large Al addition, tends to produce abnormal steel as well as fine-grained steel. This does not harden so readily on quenching as does the coarse-grained normal steel, and this, as may be seen in later discussion of "S curves," relates back to the sluggishness of austenite, and its ability, on rapid cooling, to escape transformation to ferrite and pearlite. Although the slow cooling of the carburized piece in making the examination for normality does not avoid, but is intended to produce, pearlite, it is, in a steel that has a fairly sluggish type of austenite, sufficiently rapid to avoid the breaking up of pearlite and its separation into massive cementite and ferrite.

The normality or abnormality, then, is an indication of a difference in the way austenite behaves during passage through the "critical range," the temperature range in which it breaks down into other products. This observable difference in behavior of austenite when carburized is an outward sign of differences in other behaviors that can show up in the uncarburized steels. Kerr and Eberle<sup>20</sup> conclude that normal steel, made without a large Al addition, is less readily graphitizable in elevated-temperature service. Austin and Fetzer<sup>21</sup> indicate the same thing. From most heat-treating points of view, grain size is more of a factor than normality, yet Brophy and Wyman<sup>22</sup> ascribe differences in work hardening on cold rolling and in softening on process annealing of otherwise like steels, to differences in normality.

As discussed in Chapter 11 the important point to note in respect to suitability for heat treatment is that the depth of hardening on quenching varies with grain size, *coarse grain normally corresponding to deep hardening and fine grain to shallow hardening.*



(a)  $\times 100$ .



(b)  $\times 100$ .



(a) Gas,  $\times 1000$ .



(b) Box,  $\times 1000$ .  
Nital etch. (Lauderdale, Battelle)

(a) Normal network structure. Gas-carburized 8 hr, 1700°, slow-cooled.

(b) Abnormal structure, carbide-coalesced. Box-carburized 8 hr, 1700°, slow-cooled.

Fig. 60. Normal and abnormal carburized structures in the same steel, after use of different carburizing media.

Both the grain-coarsening tendencies and the propensity toward deep or shallow hardening may vary with different heats of steel, even though the chemical compositions are closely similar and the melting and finishing practices supposedly identical.

**Hardenability.** For predicting the behavior in heat treatment by quenching and tempering, the most important test is for "*hardenability*," to appraise the response of the steel to a given rate of cooling during the quench, and particularly, to find out in what section the center will transform to an all-martensite structure. From this test it can be seen to what depth a piece of given size will harden under a quench of given intensity when it has a given initial structure and has been heated to and quenched from a given temperature. A short cut is the Jominy end-quench test, which gives the currently much-used "hardenability curves" (Chapter 11). When the "hardenability curves" and the metallographic structures at like intervals along the curves of two steels or two heats of what are nominally "the same steel" are alike, their equal suitability for a selected heat treatment, to produce the desired structure and properties, has passed the first big hurdle.

Although hardenability is primarily due to the chemical composition, and attempts are often made to calculate it from a chemical analysis, this is only an approximation. Two heats analyzing the same within very close limits may vary widely, and rather unaccountably, in hardenability. Hence testing for rather than estimating hardenability is most important.

**Transformation Rates.** If it is desired to know how much leeway exists in hardening, how much delay can be endured in getting the heated piece into the quench or, if an annealed, a normalized, or a "bainitic" structure is wanted, the rate at which the heated steel transforms to these structures at any given temperature level, then the time required for the starting of and the completion of the various transformations is useful information. Such information can be summarized into "S curves," tedious to determine, and not sufficiently informative about just what happens on continuous cooling (see pp. 317, 327). The transformation rates can be altered by varying degrees of homogenization of the heated steel, they also vary among different heats of steel, so that a single S curve on a "prototype" steel may not exactly appraise another chemically like steel. However, comparative "spot tests" of the steel in hand as to transformation rates at a few selected temperatures, combined with com-

parative hardenability tests, give an insight into how much or little the steel in hand will deviate from the behavior of the prototype.

The matter of hardenability is so important and fundamental in heat treatment that rather detailed discussion is required, but this is postponed to Chapter 11.

**Response to Tempering.** Next in importance from the heat-treatment point of view is the response to tempering of the quenched steel, the way the quench-hardened steel toughens and softens on reheating. This can be approximated from knowledge of chemical composition, and exactly determined by an actual tempering test, for which the bar used in the hardenability test may be used. Other convenient methods can be applied, the discussion of which is likewise postponed to Chapter 20.

**Necessity for Testing.** In order to find out whether a given lot of steel is suitable for heat treatment and, if suitable, to select the heat treatment most effective for the purpose in hand, tests may be applied. Likewise, the heat-treated product may also be tested.

The object of tests on the finished product is really to prove suitability for a given use. This real aim is frequently lost sight of, and, instead, the immediate aim is to meet certain arbitrary specifications which the purchaser has imposed, in the belief or hope that the specifications cover those factors on which suitability for his use depends. The specifications are usually couched in terms of certain conventional tests, so the heat treater must be acquainted with the technique of these tests.

However, such tests have severe limitations, which demand our next consideration.

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## SECTION II. TESTING

### CHAPTER 5

#### LIMITATIONS OF CONVENTIONAL TESTS

##### INDIVIDUALITY OF HEATS OF STEEL

In the preface to the AIME volume on open-hearth steel making,<sup>1</sup> Herty quotes approvingly the statement that "no two open-hearth heats are ever alike."

Individuality exists in different heats (alike as to designed chemical composition and as to processing) in that they differ in cleanliness, initial homogeneity, grain-coarsening behavior, homogeneity on heating, hardenability, and, to some degree, in response to tempering. More and more effective control is being exercised on steel making, but behavior, both as to the features making for suitability for heat treatment, and as to many of the properties in the heat-treated product, is still somewhat uncertain.

The reason is that a multiplicity of factors is involved. If everything were truly the same except the one factor we happen to be considering, say C content of unalloyed steel quenched from 1500° and tempered at 950°, we would expect to be able to plot any mechanical property against C content and get smooth curves, with perfect agreement of the data from one heat and those from another. But actually, within each lot of, say, 100, some results do deviate quite widely. To include such results the curve can no longer be drawn as a line, but must be a *band*, within which lie all the scattered points.

The heat treater does not know, without careful testing, whether a given lot of steel is on the average line or at the top or bottom of the scatter band; the user of the finished heat-treated part is likewise ignorant, without careful testing, of its position within the scatter band of a plot of whatever properties he may be demanding. And, in most cases, he cannot get a complete evaluation without making a test that destroys the part.

**The Hairpin Curve.** In either case, the average expectancy of quality and the possible deviation therefrom needs to be known to some definite degree of precision, in spite of the difficulties of sampling, inspection, and testing. The situation is quite analogous to that met by insurance companies with their actuarial tables of life expectancy. The average chance of survival for a definite number of years, of a group of 1000 men of a given age, is very accurately known, though it is not known which particular man will be the first to die, or the longest to survive. Although a multiplicity of causes will be concerned in the deaths that will occur, the number of deaths will follow a statistical law. If no factor intervenes to upset the normal expectancy, based on past experience, the deaths would plot, against time, in a characteristic symmetrical bell- or hairpin-shaped "frequency curve"; if an epidemic, on the one hand, increased, or a new remedy like penicillin, on the other, decreased the deaths, the curve would be distorted.

Examples of (a) the usual expectancy when no unusual variable comes into play, and (b) the expectancy where the normal curve is distorted by the effects of (in this case three) unusual variables, were given by Saklatwalla and Chandler.<sup>2</sup> They cite the application of probability plotting to a foundry engaged in casting flywheels, where a certain Brinell specification had to be met. Their Brinell readings for one month are plotted in a probability curve in Fig. 61a. This shows that their composition and practice are nearly right for meeting the specifications. The only fault would be that the iron had somewhat too wide limits of hardness. However, conditions are not bad; the percentage of rejects is about eight. Figure 61b shows the results for the following month, plotted in the same way. Here it may be seen that the probability curve no longer has a regular shape. This indicates that the probability curve is a summation of several different curves indicated by the dotted lines. These results show that there are certain influences, apparently three, which are tending toward the production of softer iron. The plot does not tell *what* happened, merely that *something* happened. These influences reflect some changes in foundry procedure, and a knowledge of what has been happening in the foundry would enable these changes to be corrected. A mere average of the hardness results would not tell anything, since the average hardness of the results is the same within two points in Fig. 61a as in Fig. 61b.

Study of the frequency curves often brings out the fact that a set of observations from which we are hoping to arrive at the normal

behavior and the range in behavior may be too limited in number to give true indications of either. Likewise we may be sampling or testing too small a proportion of our output to keep from blinding ourselves to the possibility of defective material being shipped and to the risks we are taking of having that defective material show up in service, come back, and injure our reputation. On the other hand, we may be making many needless tests. Statistical methods, properly applied and used from a common-sense rather

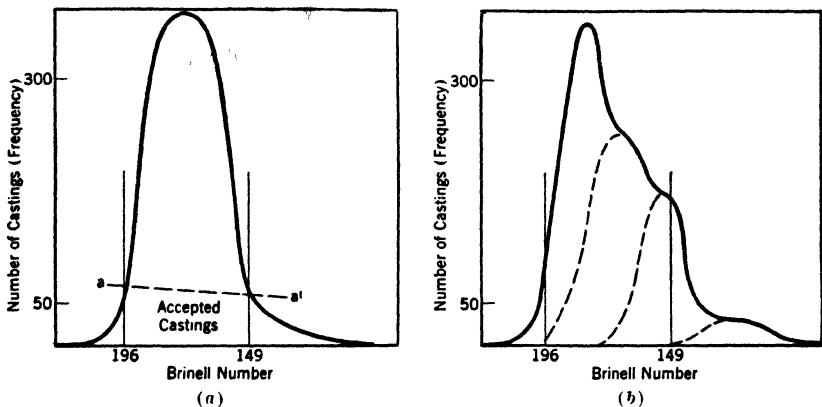


FIG. 61. Probability curves of Brinell hardness of cast iron. (Saklatwalla and Chandler<sup>2</sup>)

than too highly mathematical a viewpoint, can clarify these problems.

Statistical methods are valuable for determining the suitability of an alloy for meeting certain specifications. From probability curves we can find not only what proportion of our output is meeting specifications but also whether its composition and treatment have been right to make the average come in the middle of the specification range. Also, a lop-sided probability curve will show that there are certain influences present whose correction is necessary in order to obtain more consistent results.

If only nondestructive tests, such as hardness, are involved in meeting specifications, we can, of course, test each piece in order to be sure that none of the output will fail to meet the requirements. When destructive tests, such as the tensile test, must be performed on the object itself, probability curves indicate what percentage of the output will meet specifications and how technique should be varied the better to meet specifications.

A sensible question is, "Why not determine the properties once and for all, and then make each succeeding piece in just the same way, so they will be alike and there will be no necessity for testing?"

This method applies pretty well to 4-min eggs, provided the eggs are fresh. It is a bit more difficult to apply to biscuits or popovers, and still more difficult to apply to steel, with all the variables that affect steel making and steel treating. Nevertheless, it is a good mark to aim at.

#### STATISTICAL CONTROL

Much effort is being devoted toward hitting this mark, and considerable know-how, applicable to any manufacturing process, is being acquired and systematized, by utilizing probability principles.

In a given practice there is a normal amount of scatter. If a running account is kept of results from routine inspection, once the degree of variation is known that occurs from the chance combination of several variables, none of which has a major effect by itself, a continued variation *outside* the range of normal scatter indicates that it is not mere chance that is acting, but that some important factor has been altered. Probability plotting makes this evident, but does not reveal what that factor is. To find the factor, one re-examines the raw materials and all the steps in processing, with a Sherlock Holmes's attitude of mind. The day-by-day plotting puts the observer on guard and lets him get in his detective work before many more crimes have been committed.

Untangling the cause, amid the conglomeration of minor variables affecting plant production, may be difficult, and so, if the intruding cause is not spotted promptly, the problem may have to go to the laboratory for studies in which all other factors than the one suspected are most rigorously controlled, and known variations in the suspected one are introduced.

Keeping the day-by-day record, and applying it, is "statistical control," a method which finds increasing application. The assurance it gives, when the results fall within the normal scatter, that the product is normal, relieves one of the necessity of making as many tests on the final product as would be necessary were there no such assurance; and, conversely, imposes the necessity of more rigorous inspection of the product made during a period when the probability plot says that something or other is "out of control." The technique of statistical control has been quite highly developed. During the war, it saved many man-hours of inspection and

testing, brought otherwise unsuspected troubles promptly to light, thus allowing them to be remedied, and made feasible the production of many more acceptable guns and much more ammunition than was the case when statistical control was not attempted.

An explanation of the mechanics of the method, applied to a metallurgical problem, is given by Motley.<sup>3</sup> A great deal of literature on statistical control has been published in recent years, ranging from articles and books with a highly mathematical approach to others, such as an article by Cutter,<sup>4</sup> which point out that good horse sense plus a willingness to use the tools provided by the mathematician makes the method usable without requiring that the user be able to derive the basic mathematical formulae.

Many of the publications in this field relate to maintenance of the specified dimensions on products of the machine shop. In such cases, the causes for deviations from what is aimed at are usually quite obvious, once the deviations have been noted. In metallurgy, locating the cause is not always quite so simple, and finding it requires metallurgical background and experience. The combination of a quality-control man to detect deviations immediately as they occur and a metallurgist to ferret out and remove the changed conditions that led to the deviations is an effective one.

Metallurgical cases are cited by Rogers,<sup>5</sup> Butler,<sup>6</sup> Olds,<sup>7</sup> Westerman and Freeman,<sup>8</sup> and others,<sup>9-11</sup> in accessible publications. Pertinent literature is shown in their bibliographies, and some of the general articles and books dealing with the general problem are listed here.<sup>12, 13</sup>

**Inspection.** Probability methods have been put to use in proving that severity of inspection, resulting in more rejection, increased while the quality of the product did not actually change.

The futility of reliance on the result of a single test for appraising a whole heat or a large lot and the necessity for expressing trends of properties by plotting bands instead of lines clearly emerges from such studies.

One of the earliest studies of the scatter of results in a heat-treated steel was made by Janitsky<sup>14</sup> nearly 20 years ago, long before the days of modern grain size and hardenability control, on 6130 Cr-V steel. The presence of V doubtless gave a fine-grained steel, so that the spread of results was probably smaller than would have been met in steels without V. Four steels, to meet the specifications for 6130, were made by four steel companies, were rolled to 1-in. rounds, and were distributed to 30 laboratories. Later 3130 was similarly

distributed. The data for chemical analysis and for ordinary tensile properties of specimens water-quenched and tempered at three

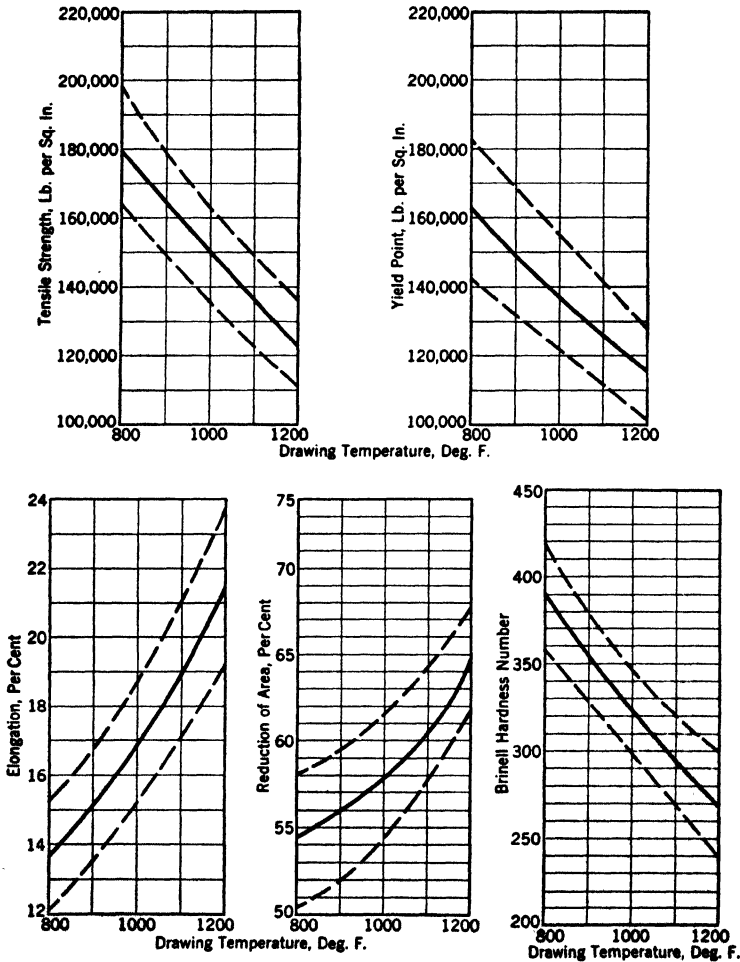


Fig. 62. Summary of mechanical properties of Cr-V steel 6130. Maximum and minimum values are represented by dashed lines and averages by solid lines. This is the "scatter-band" type of plot. (Janitsky <sup>14</sup>)

temperatures were collected to show the spread, and were plotted in scatter bands and frequency curves, shown in Figs. 62-67.

**Property Charts.** As a result of the information thus brought out, the SAE abandoned the practice of publishing in its handbook "property charts" plotted as lines, recognizing that they were mis-

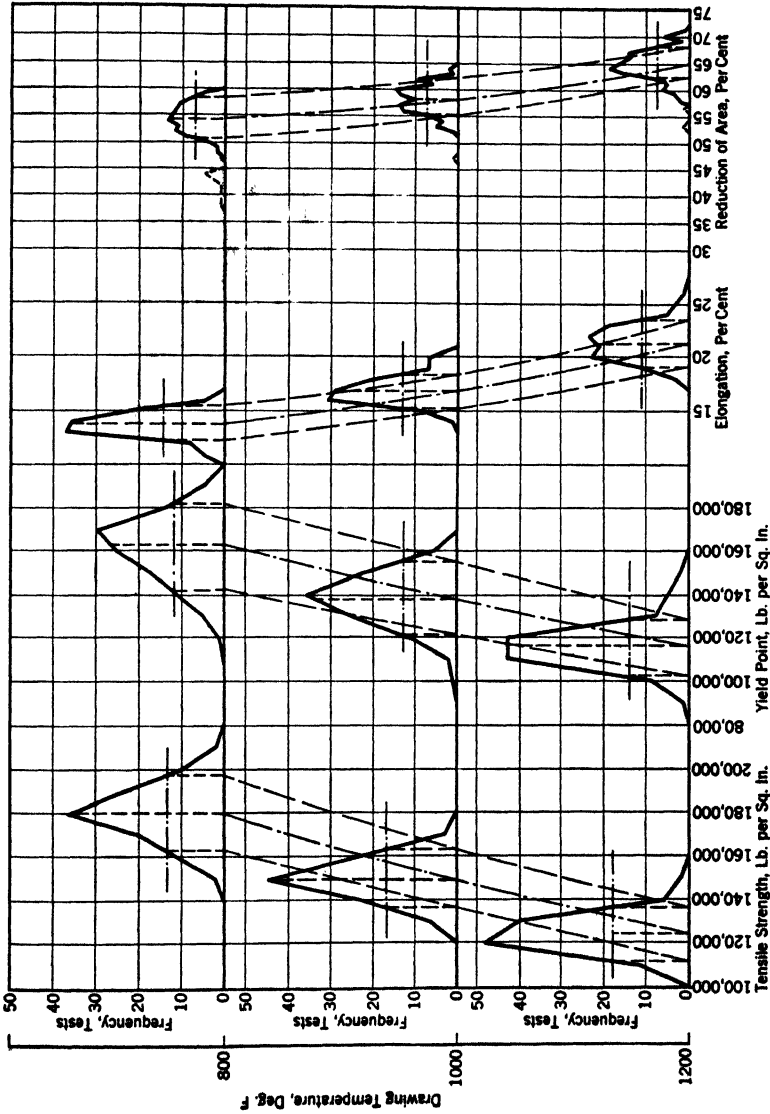


Fig. 63. Mechanical properties of Cr-V steel 6130 drawn at 800° (right), 1000° (center), and 1200° (left), plotted as frequency curves. (Janitsky<sup>14</sup>)



leading. A further complication was involved, in that, since readers were likely to consider such published values as minimum values, some steel makers' property charts showed low values for both strength and ductility (which properties really have an inverse

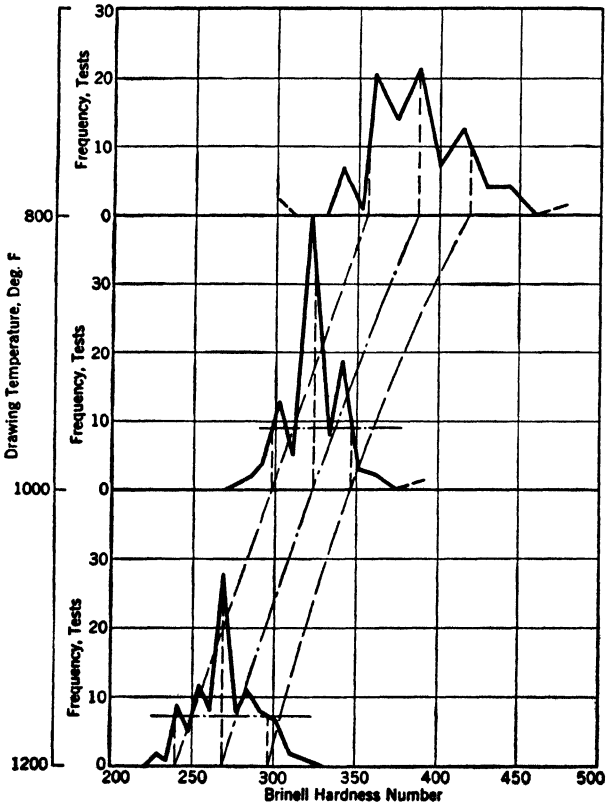


FIG. 64. Frequency curves of hardness tests of Cr-V steel 6130 drawn at 800°, 1000°, and 1200°. (Janitsky<sup>14</sup>)

relationship) and, in their conservatism, distorted the truth. Other steel makers gave average values, and considerable differences thus appeared among published data for the same steels. These discrepancies are accounted for when the plots are given as bands, as in Figs. 62 and 67.

Moreover, as more information was collected, it became evident that, as long as the steels compared had all been fully hardened to martensite and tempered back to the same hardness, or tensile strength, instead of at the same temperatures, the property charts

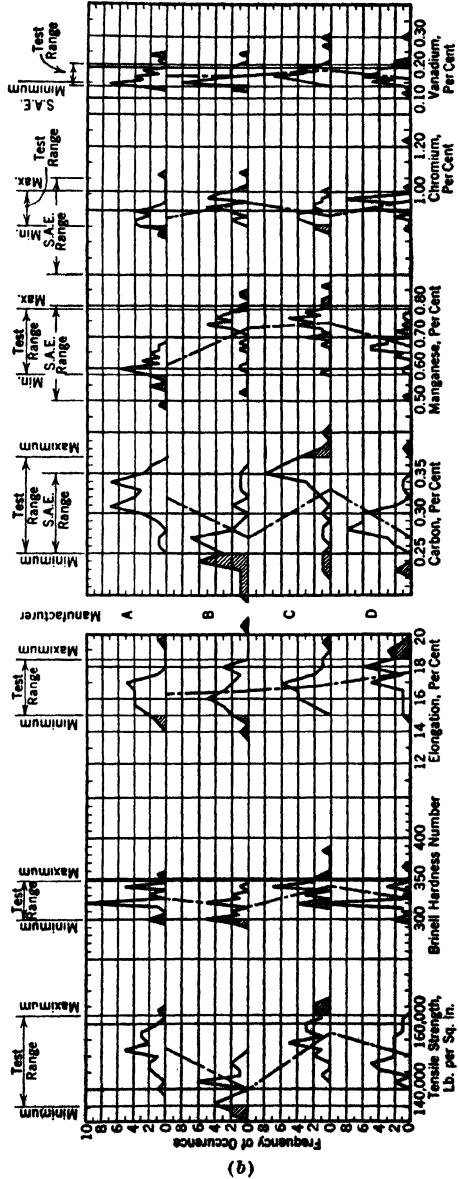
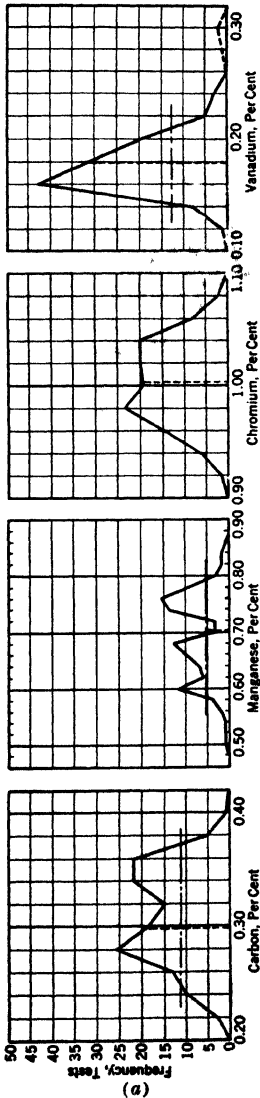


FIG. 65a. Frequency curves for chemical analyses of the four steels whose properties have been lumped in Figs. 62-64.

FIG. 65b. Frequency curves for analyses and for corresponding mechanical properties, of the four steels, separately. These four Cr-V 6130 steels were water-quenched from 1600°. Fig. 65b shows data only for the 1000° draw. (Janitsky 14)

all coincided closely in respect to those properties usually charted. This coincidence is brought out in Figs. 68-70. A couple of property charts, shown in the previous edition, are reproduced as Figs.

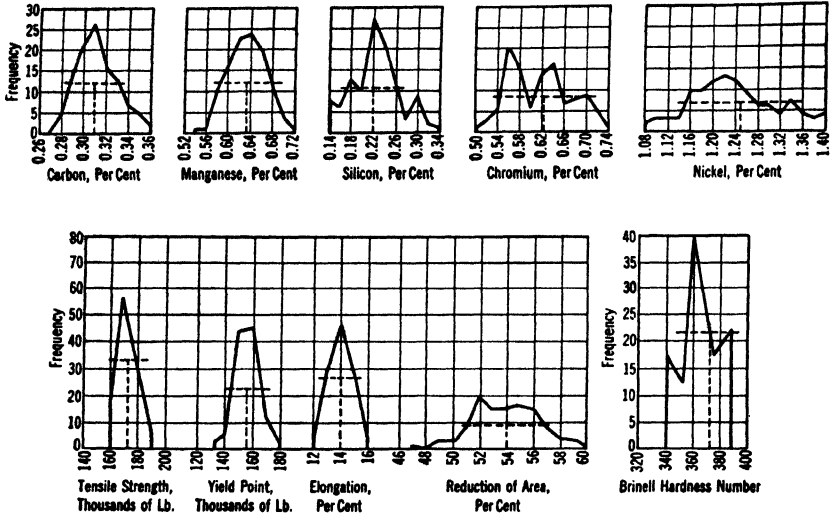


Fig. 66. Frequency charts for Ni-Cr 3130 steel. Chemical composition at top: mechanical properties, water-quenched from 1525°, drawn at 800°, at bottom. (Janitsky<sup>14</sup>)

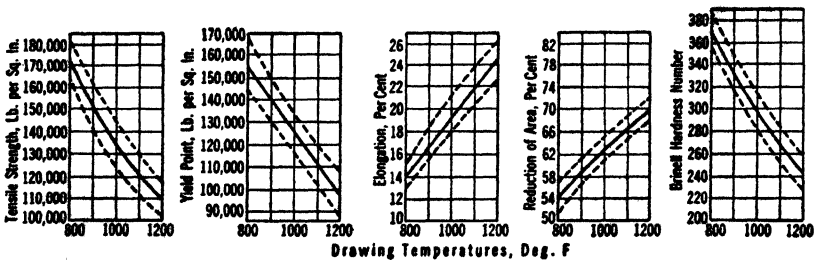


Fig. 67. Scatter-band plots for the Ni-Cr 3130 steel at draw temperatures from 800° to 1200°. (Janitsky<sup>14</sup>)

71 and 72. If the data from these plots are entered in the plots of Figs. 68-70, it will be found that the latter figures are adequate to express the information for practical purposes. Patton<sup>15</sup> plotted data from 180 heats of SAE, N.E., and other steels, which all fit these same plots.

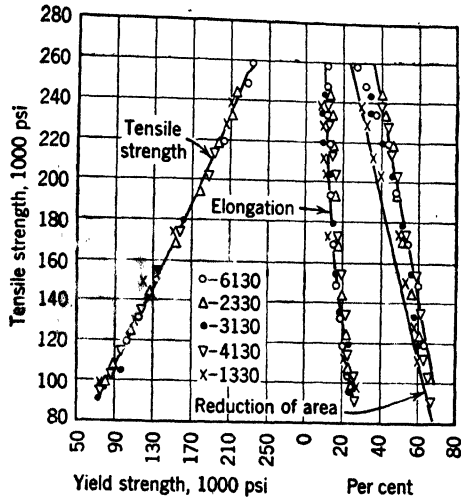


FIG. 68. Tensile properties of water-quenched and tempered SAE steels. (Janitsky and Baeyertz, *ASM Metals Handbook*)

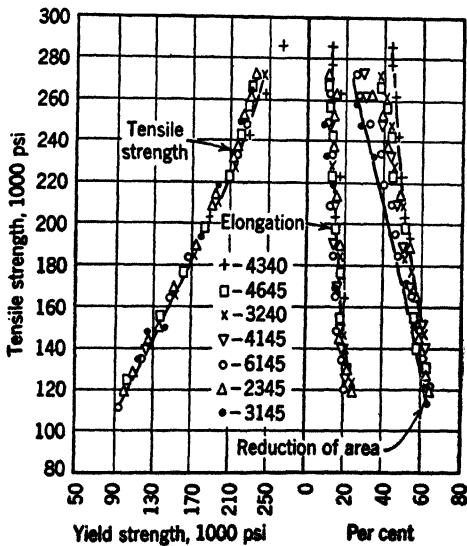


FIG. 69. Tensile properties of oil-quenched and tempered SAE steels. (Janitsky and Baeyertz, *ASM Metals Handbook*)

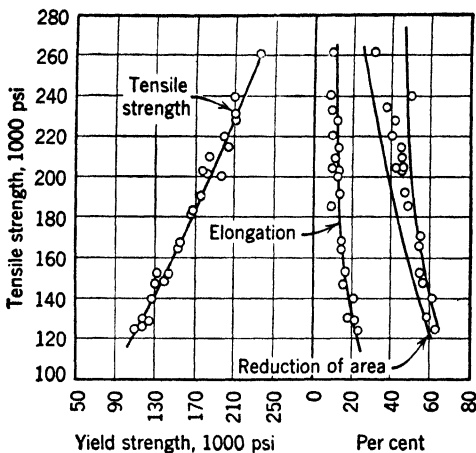


Fig. 70. Tensile properties of quenched and tempered N.E. steels. (Data from references 15-19)

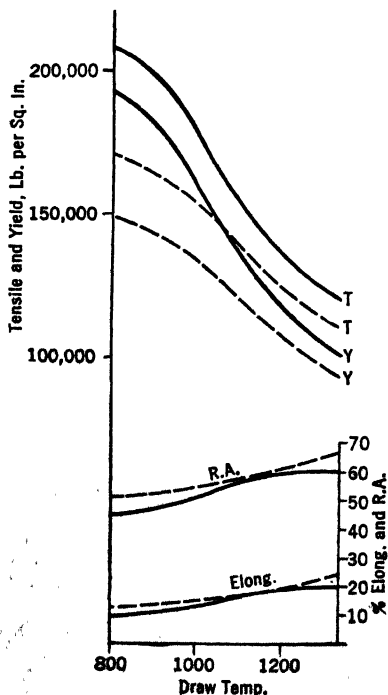


Fig. 71. Properties of 4130. 0.30 C, 0.20 Si, 0.65 Mn, 0.65 Cr, 0.20% Mo. Water-quenched from 1550°, solid lines. Oil-quenched from 1600°, dashed lines. (Bethlehem Alloy Steels)

In *Republic Alloy Steels* quite variant results are given, the properties for 4130 water-quenched being not far from those here shown for oil quenching.

That is, instead of revealing differences in the properties the engineer is interested in, the property charts merely told what tempering temperature was required to produce a given tempered structure and the particular combination of properties that accompanies it! Where true differences in behavior exist, as a result of chemical composi-

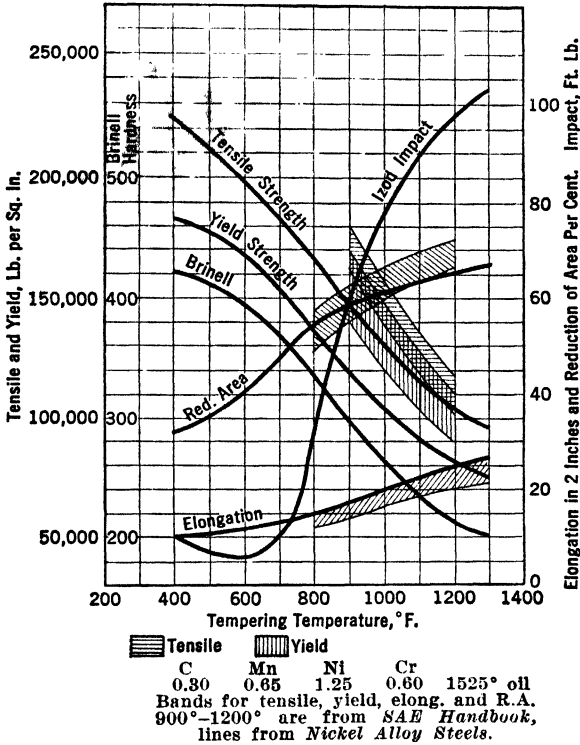


Fig. 72. Properties of 3130, in small sections, about 1 in., oil-quenched and drawn as shown.

tion and methods used in finishing the heat of steel, other properties and other modes of plotting are far more revealing.

Even a slight acquaintance with the probability point of view, which can be gained without going into difficult mathematics, is so illuminating to those who make tests or draw conclusions from test data that it may well be classed as an essential part of the information of the modern testing engineer as is an understanding of the fact that a tensile machine that is out of calibration will not give reliable results.

**Quality.** Recognition of the natural variability in results, so clearly brought out by use of probability methods, makes it necessary to view the question of steel quality and uniformity in the light of actual service requirements. Design may call for use of the same steel, and the same strength, in a farmer's cultivator and in an aircraft engine, but a failure of the cultivator is just an annoyance, whereas that of the aircraft engine may be fatal. From the cold-blooded cost angle, the cultivator failure means a small loss of time and the cost of a repair part, that of the engine may wipe out an exceedingly costly plane, and so we can pay more for assurance of uniform quality in the steel in the latter case.

This has been well stated by Archer,<sup>1</sup> as follows:

#### QUALITY

No steel is equally well suited for all purposes. This situation leads to the concept that quality can be defined only in terms of the intended use. The steel of the best quality is the one that best meets the purpose for which it is made. Long experience of both steel producers and consumers indicates that the time and expense required to produce properties better than those required for a specific product are wasted. If a large proportion of the steel produced for a certain product must be very much above the requirements in order to insure that none falls below the standards set for this product, the cost will probably be too high. Therefore, perfect conformity to inspection standards is likely to indicate unwarranted expense. The importance of this factor is expressed in a statement by B. M. Larsen: "The optimum level of quality is indicated by 1 per cent rejections." The 1 per cent figure is purely arbitrary and may be changed by variations in product or practice. The important point is that it is usually more economical, from the standpoint of both the steel producer and steel consumer, to permit a certain amount of defective material. In summary, quality is made up of consistently reproducible uniformity at a level determined by application and cost. This has been expressed well by Walther Mathesius in the following definition:

"From a practical and economic viewpoint, steel quality is that group of properties essential to the constant production of finished goods of merit measured in terms of satisfactory service rendered at the lowest ultimate cost. In this sense it is an ever changing and a competitive standard."

In some instances the requirements for the final product into which a heat of steel is going to be made are clearly known; in others they are not. For example the service conditions and requirements for rails and wheels or structural members are comparatively well known; but the final application of, and thus the requirements for, the products into which an order of merchant bar or jobbers sheets is fabricated are often unknown. This means that all rails can be judged by about the same standard. A sheet, however, that is of first quality for the surface and flatness requirements of furniture stock would be of very poor quality if used for deep drawing.

This topic is discussed by Parker.<sup>16</sup>

**Limitations of Testing for Mere Similarity.** The designer sets certain limits of acceptance for the steel parts he is going to build into something to serve a definite purpose; his purpose being to secure either, or both, amenability to processing or freedom from failure in the final "something." He tries to accomplish his purpose by writing a specification. A definite specification demands that the product delivered to him must pass certain tests.

All too seldom do the specified tests really measure the properties truly needed for service; far more often they represent some few properties determined by some conventional test to exist in the steel or other material that he used before and found to perform acceptably. He wants more like it, and, for the moment, is not concerned with how much cost he could avoid, how much weight he could save, or what other improvement would be possible, were another material chosen, another set of specifications drawn, or a design modification made to allow use of another material with different properties.

He may deceive himself into thinking that the conventional tests he says must be met are true service requirements when they are not, or he may recognize that those tests are not the crucial ones; but, for lack of knowledge as to how to carry out truly crucial tests, he adopts the noncrucial conventional tests in the expectation that similarity of unmeasured properties will come automatically with similarity of those that are measured. The expectation that these tests for similarity prove utility is not always fulfilled, and so the steel maker, the metallurgist, and the steel treater may well quiz the designer about the service conditions so that they may provide material which meets the unspoken requirements. Were those requirements really understood, the specification would often read quite differently, would allow wide variations in properties that do not count, and would demand strict control over those that do.

**Variety of Properties.** Boulton<sup>17</sup> gives a list in Table 4 of some of the properties that may be of interest to the aircraft engineer. This emphasizes the fact that the ultimate use must be considered in selecting the tests for evaluation of suitability.

This is becoming more widely understood, and many special tests are being made to evaluate properties not revealed by the conventional tests. The special tests may be applied as acceptance tests, but many of them take too much time or effort to make them popular. Conversely, some simple tests that bear no direct relation to



TABLE 4

CHECK LIST FOR CHARACTERISTICS OF A METAL

Purpose of this list is to present reasonably complete listing of all properties of a metal so that, when a specific metal is under consideration, those desired properties pertaining to it may be readily selected and their priority determined. The list is of assistance in assuring that no properties are overlooked.

General	Physical Characteristics	Mechanical Properties	Fabricating Characteristics	Economic Factors
Density	Tension modulus of elasticity	I. Machinability: Boring and turning Machining speed Material and shape of cutting tool Lubricant or coolant Fire hazard Qualitative suitability for: Routing Sawing Shearing Drilling Turning Milling	III. Formability at elevated and room temperature: Heat-treating charac- teristics Annealing procedure Tempering procedure Quenching procedure Sensitivity to variation in aging characteristics	I. Cost in different forms II. Availability: Single or Multiple source III. Availability in different forms: Sheet Bar Bare or clad Wire Tubing Extrusion Impact extrusion Pressing Forging Casting: Sand Centrifugal Die casting Powder Sintered powder
Thermal coefficient of expansion	Compression modulus of elas- ticity	II. Formability in: Annealed and tem- pered states True stress-strain curve Apparent stress ver- sus local strain curve	IV. Weldability: Arc welding Torch welding Type of welding rod Cracking tendency Helarc welding Atomic hydrogen weld- ing	IV. Size limitations in dif- ferent forms: Weight Gage Length Width Future tolerances
Thermal conductivity	Complete stress-strain curve both tension and compression	Elongation versus gage length from zero to 100 in. Standard hydro-press specimen test Characteristics in: Shrinking Drawing Stretching Dimpling Jogging Bending Uniformity of charac- teristics	V. Brazing and soldering: Flux and rod VI. Torch cutting Quality of finish (smoothness) possible VII. Protective coating neces- sary for: Storage Processing Service Packing Adding Point of use Sensitivity to contami- nation	V. Availability as to time: Present Future
Position in electromotive series	Shear modulus of elasticity	Hardness	VIII. Protective coating neces- sary for: Storage Processing Service Packing Adding Point of use Sensitivity to contami- nation	VI. Availability as to quan- tity: Future tolerances
Corrosion characteristics in: (ASTM committee B-3)	Poisson's ratio	Impact properties: Endurance limit Notch sensitivity Impact properties: Effect of low temperature Above strength properties: (a) At elevated temperatures and for long holding times at temperature (b) At room temperature after exposure to elevated temperatures	VIII. Protective coating neces- sary for: Storage Processing Service Packing Adding Point of use Sensitivity to contami- nation	VII. Size tolerances for all forms
Atmosphere	Compression yield values	Uniformity of charac- teristics		
Salt water	Tension yield values			
Acids	Design with and across grain			
Hot gases	Shear ultimate			
Rapidity of corrosion	Bearing ultimate			
Susceptibility to stress cor- rosion	Tangent modulus curves in com- pression—with and across grain			
Effect of high temperatures on corrosion resistance	Fatigue properties: S-N curve Endurance limit Notch sensitivity Impact properties: Effect of low temperature Above strength properties: (a) At elevated temperatures and for long holding times at temperature (b) At room temperature after exposure to elevated temperatures			
Maximum temperature not af- fecting strength character- istics	Effect of high temperatures on corrosion resistance			
Effect on strength after exposure to higher temperatures	Effect of high temperatures on corrosion resistance			
Fire resistance	Effect of high temperatures on corrosion resistance			
Hydraulic permeability in differ- ent forms	Effect of high temperatures on corrosion resistance			
Creep characteristics versus tem- perature	Effect of high temperatures on corrosion resistance			
Melting point	Effect of high temperatures on corrosion resistance			
Weldability	Effect of high temperatures on corrosion resistance			
Effect of cold working	Effect of high temperatures on corrosion resistance			
Working quantities:	Effect of high temperatures on corrosion resistance			
Introduced by heat treatment	Effect of high temperatures on corrosion resistance			
Given by plating	Effect of high temperatures on corrosion resistance			
Ballistic properties	Effect of high temperatures on corrosion resistance			
Damping coefficient	Effect of high temperatures on corrosion resistance			
Crystal-structure type	Effect of high temperatures on corrosion resistance			
Magnetic properties (Curie point)	Effect of high temperatures on corrosion resistance			
Chemical composition	Effect of high temperatures on corrosion resistance			

the evaluation of the properties really required by service are frequently made just to give a more complete survey of the degree of likeness or difference between the current lot and a previous lot. Such additional information is welcome, but, when the assumption is made, without proof, that a difference has a causal relationship to behavior in service, and specifications are drawn on that assumption, material that service tests would show to be entirely satisfactory may be excluded. For example, if, with the idea of increasing the certainty of similarity, the designer multiplies the tests, adding, say, low-temperature single-blow notched-bar impact tests, when the actual use is to be repeated stress at high temperature, the path not only is devious, but also does not necessarily lead in the right direction.

**Plastic Behavior.** An oft-cited and highly pertinent example of such a case is the Ford cast crankshaft. When forged-steel crankshafts were used, high ductility and high impact resistance were specified, really because these properties were characteristics of the forged steels used, after proper heat treatment. But the designer was inclined to feel that the toughness specifications were imposed in order to prevent a shattering failure of the crankshaft, to allow it to bend without breaking, as an insurance factor. He closed his eyes to the fact that a bent crankshaft means a ruined engine, just as much as does a broken one, so that the "insurance" was illusory.

Experience with bent crankshafts led to increasing the section and the stiffness of the forged crankshaft. Finally the Ford designers sensed that service demanded stiffness, yield strength, fatigue resistance, and wear resistance. It did not involve impact, nor could it use ability for deformation. By shaking out the unnecessary requirements and imposing only those needed, they were emboldened to apply the cast material, meeting all the needed requirements but having almost zero ductility and close to zero notched-bar impact resistance.

Parts liable to be subjected in service to shock loading or to infrequent static overloads of high intensity are necessarily made of something that is not "glass-brittle," but has some ability for plastic deformation. Cast-iron valves and fittings, though they do have one type of toughness, are not considered suitable for use on warships, for a shock transmitted through the ship from a torpedo or other hit somewhere on the ship may shatter them. Ability for slight local plastic deformation is called for in a riveted structure in order that, by such actions, the stress, which originally was high

at certain local points because of unavoidable inaccuracy in alignment of the parts joined, may be distributed.

The fit of a bolt and nut is improved by slight plastic flow, occurring when they are screwed together under load.

Under repeated stress, it is desirable that a locality of high stress concentration, such as the base of a sharp notch, be able to deform slightly, become more rounded, and thus reduce the stress concentration, instead of forming and propagating a crack. Such ability is termed "crackless plasticity."

High static ductility is commonly assumed to connote freedom from "glass-brittleness" in impact and capability for redistribution of static stress, and to indicate "crackless plasticity," or resistance to repeated stress, but the correlation is only qualitative. As a rough line of demarcation between "brittle" and "tough" materials, Marin<sup>18</sup> takes 5% elongation in a static tensile test.

**How Much Ductility Is Enough?** Miller has been cited by Epstein<sup>19</sup> as believing that 5% elongation about a rivet hole would be quite adequate for riveted structural-steel members. Structural steel ordinarily has 20 to 25% elongation. Moore<sup>20</sup> considers that deformation only of the order of 1% over a minute area may accompany satisfactory "crackless plasticity."

Koegler and Schmitt<sup>21</sup> studied aluminum-alloy aircraft-wing construction and showed that 1½% tensile elongation, utilized as permanent deformation, was excessive, for it was sufficient to ruin the aerodynamic efficiency of the wing. They comment that the 10 to 14% elongation inherent in the material can never be utilized in service of the wing.

It is notorious that neither static notch sensitivity, as indicated by single-blow tests, nor notch sensitivity under repeated stress, bears any direct relationship to static ductility. They vary with the structure and with the kind of alloy being considered.

While present-day gun steels have higher yield strength and greater ductility, Dickson<sup>22</sup> gave, for a usable cast autofrettaged big gun, in finished condition, representative values of 115,000 psi tensile, 90,000 psi yield, 5 to 11% elongation, 36 to 48% reduction of area. If 5% is usable in such service, then the ductility figures of most specifications do not relate to the engineering requirements of the service but rather to that level of ductility that accompanies the required yield strength in normally processed material.

The primary role of tensile ductility measurements is to indicate soundness and cleanliness of the material, by comparison with nor-

mal values for the same type of material in the same structural condition and at the same strength and hardness level. It is in this comparative sense that they are cited in our general discussion of the properties and behavior of steel.

Knowlton<sup>23</sup> puts it: "There is a useful limit of plasticity depending on the design of the article and the type of service it must withstand. A higher degree of plasticity would not increase the merit of the steel as far as that particular application is concerned."

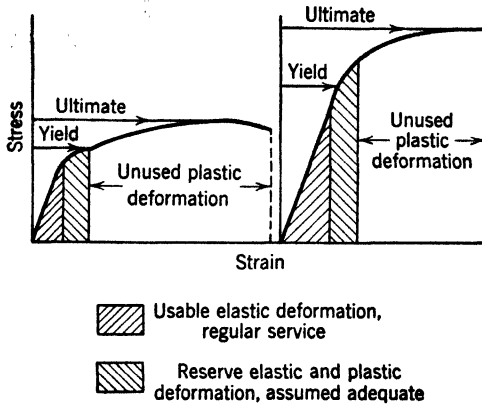


Fig. 73. Diagrams for a low-strength and a high-strength steel. The unused plastic deformation is not needed in normal service, even that requiring slight plastic adjustment.

The situation as to actual and usable plastic deformation is schematically shown in Fig. 73. This and other matters relating to interpretation of tests have been discussed.<sup>24</sup>

Because deformation on a very local scale is involved in "crackless plasticity," some authorities favor using reduction of area as the criterion rather than elongation and tend to evaluate quality on the basis of a few per cent difference in R.A. The R.A. percentage figures overemphasize the differences. Even 5% R.A. means that an 0.505-in. section has thinned down to 0.492 in. and has shown considerable ability for redistribution of stress. At 50% R.A. the final diameter is 0.357 in.; at 55%, 0.339 in. It is difficult to visualize a use in which 50% is not enough and in which 55% would display any real advantage. Figures 68-70 indicate a considerable scatter in R.A. values. Stainless steel, manganese bronze and various other alloys deform chiefly through general extension, rather than by local thinning down.

There is no magic in an elongation or R.A. figure in itself, nor is it any sure measure of ability for plastic deformation under service conditions. Specifications based on the conventional tensile test are not certainly adequate to evaluate toughness that may be desired for accident insurance in a finished part or formability that may be desired in processing. A new untried alloy might have 50%, 5%, or even 2% elongation, and be equally well suited for a particular service. It should be neither accepted on the basis of high ductility nor rejected on the basis of low ductility.

Badger and Sweeney<sup>25</sup> bring up this point in relation to special alloys for new gas-turbine designs. Very superior alloys are available, but they have less than 5% elongation. The question whether the service really requires more or whether the requirement of more is merely due to the conservatism of the designer is no merely academic matter. It will have to be settled by actual test and experience, not by anybody's say-so.

Aitchison<sup>26</sup> suggests critical examination of ductility requirements to give the metallurgist more scope for production of high-strength alloys for aircraft use.

**Specifications.** It is often necessary to control the design and the service conditions to accommodate some deficiency in a material, condoning that deficiency because of otherwise outstanding suitability.

Glass is fragile, but we use it with care. The man who dropped and broke his glasses and in disgust ordered his next pair ground to his prescription in stainless steel, didn't get a very good pair.

Such points of view emphasize that it is pertinent to question any specification that has been based on the principle of likeness to something used before, rather than on actual service requirements. It is also pertinent to find out whether the specification has not omitted mention of some attribute really necessary in service, which might be supplied by proper choice of material and processing; in our case, of steel and its heat treatment, so that the metallurgist and heat treater may supply what the user really wants, in spite of his failure to specify it.

Examples of unrealistic specifications are not hard to find. A specification (AREA 56,542)\* for steel for railway bridges, drawn before the days of high-yield-strength mild-alloy steels, states "no steel should be considered satisfactory for bridge construction if its

\* American Railway Engineering Association.

yield point exceeds 70% of its ultimate strength." Experience with modern mild-alloy steels having yield ratios of 70.5 to 80% amply shows that these steels are suitable for bridge construction. In fact, an English producer of steel of this type has made prefabricated bridges of such steel for shipment to India and points out that the lower weight of an adequately strong bridge made of high-yield-strength steel led to great savings in freight costs. The Bailey bridges, so useful during the war, were necessarily made from high-yield-strength steel, for ease in handling.

A Federal specification for ferrotitanium once included a 1% upper limit for Si, for no reason save that it made it sound like a well-drawn specification and the producers said they could meet it when the specification was first drawn. Later, Ti ores higher in Si were being used, the product ran over the 1% Si limit, and was being rejected on chemical analysis. Since the alloy was for use in cast steel, the added trace of Si introduced by it was of no importance. Rejection of the available ferro hampered production of cast steel needed by the Government. The producers clamored for a modification, which was granted. This difficulty was the fault of the producers who had allowed a meaningless clause to be inserted without opposition.

Specifications ought to carry footnotes referring to records of experimental facts to justify the restrictions imposed.

The specification of the American Welding Society and the American Society for Testing Materials (A233-40T), for arc-welding electrodes calls for building up a deposit of weld metal from which a standard tensile specimen is machined and which must meet certain requirements as to ductility. The deposit is built up in several passes, each not over  $\frac{1}{8}$  in. thick. After each pass that layer of deposit is allowed to cool for a specified time, then the weld is immersed in boiling water for 5 min, another layer deposited and so on. No explanation is given for this deviation from the practice of actual welding. Presumably, the idea is to facilitate the escape of hydrogen (the retention of which would lower the ductility) with the assumption that hydrogen will, in time, escape from an actual weld, without its presence having done any harm, an assumption the correctness of which is doubtful. Thus the specification imposes an artificial set of conditions which may or may not make sense, but which certainly do not do so without explanation.

Specifications for steel to be cold-worked, bent, flanged, deep-drawn, etc., are likewise often phrased in terms of the conventional

static test, with especial attention to ductility and contraction in area. The yield strength after cold working may be important, but is not necessarily evaluated by the yield strength measured on the material before cold working. While the ductility is specified, what is actually wanted is formability. Formability may involve stretching, or it may require a type of upsetting, or many combinations of these and other types of deformation.

It is notorious that the conventional tensile ductility test often fails to differentiate between steels of good and of poor behavior in a particular forming operation. Boulton<sup>16</sup> pointed this out for Al alloys; the same thing is true of steel.

Conventional fatigue and notched-bar impact tests also have their limitations, which can best be discussed in connection with those tests in the next chapters.

As an example that similarity in mechanical tests may fail to reveal amenability to processing, two steels indistinguishable as to mechanical properties before or after heat treatment, when the C level is held constant, may differ widely in their propensity toward decarburization in heating for rolling or quenching.

#### DIRECTIONAL PROPERTIES

Many of the parts handled by the steel treater are of small cross section, made from rolled or forged stock in which hot working has oriented the elongated grains, the ferrite bands or carbide stringers, and the inclusions that are plastic at hot-working temperature, or has strung out the nonplastic inclusions, all in the direction of working. Unless a plate is cross-rolled or a forging worked equally from all sides, the structure has a laminated appearance. Some of this structure tends to persist in spite of heat treatment.

Because of the dimensions of much of such stock, too small from which to take conventional specimens (though plate can be sampled in two directions) the tensile test is ordinarily made only on specimens taken in the longitudinal direction. The tensile and yield strengths seldom differ greatly between the longitudinal and transverse directions, but the ductility may vary widely, and characteristically shows a much greater spread in transverse specimens or those oriented within some 30° of it. This spread is on the low side; the frequency curves "tail out" on the low side, so that normal material may give some relatively very low transverse ductility values, a condition not hinted at by the longitudinal tests. This is another

reason why a difference of a few per cent in longitudinal reduction of area is a poor criterion of quality. Service stresses are not necessarily all applied purely in the longitudinal direction. When cold forming is required, it is generally found that bending or other de-

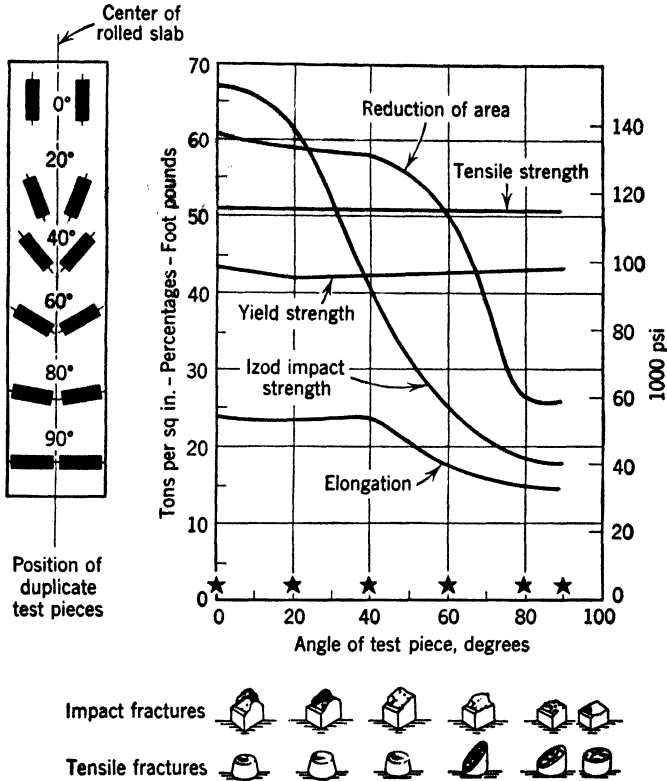


FIG. 74. Effect of angle of test piece relative to direction of fiber on mechanical properties of steel. (Brearley)

formations in a direction tending to split the piece along the laminations cannot be so severe without failure, as when done in the other direction.

Thus some of the forged crankshafts previously referred to may easily have had very low transverse ductility in spite of high longitudinal ductility, without the user ever having been aware of it, in the absence of transverse data. It is fortunate that ductility plays so minor a part in their actual service. Directionality is a characteristic of the heat of steel, developed by the deformation of mechanical work. It is affected by degree of reduction in working, as



well as by the amount of softening by tempering a hardened steel. The same heat of steel with the same mechanical reduction may, in a transverse specimen, show reductions of area only half that of a longitudinal specimen when tempered to 200,000 tensile, but 80% of the longitudinal value at 100,000 tensile. Cross rolling of plate tends to equalize the longitudinal and transverse values, but the value in the third direction, that of the plate thickness, may still be very low.

Brearely long ago gave data on directional properties, so illuminating that his figure has been reproduced many times, and is again here reproduced as Fig. 74.

**Low Through-Plate Ductility.** Even though the longitudinal and transverse ductility in rolled plates may not always vary so widely as Brearely's data indicate, surprisingly low ductility may be present in the direction normal to the plate. Thus, if one takes the trouble to examine subsize specimens taken in the direction normal to the plate, he is likely to find, in 1-in. plate of some high-strength mild-alloy steels, as rolled, results like the following:

	<i>Tensile,</i> <i>Psi</i>	<i>Yield,</i> <i>Psi</i>	<i>Elongation,</i> <i>%</i>	<i>R.A.,</i> <i>%</i>
Longitudinal	65,000-75,000	45,000-50,000	30	65
Transverse	65,000-75,000	45,000-50,000	20	55
Normal	65,000-75,000	45,000-50,000	2 to 4	5 to 15

Directional properties of certain other hot- and cold-rolled steels are discussed on p. 240.

This huge variation in ductility in wrought steels, with direction, is reduced in fine-grained quenched and tempered steels, but a considerable variation is to be expected. When the properties in other directions are ignored, it is straining at a gnat and swallowing a camel to pay much heed to the scatter of longitudinal R.A. values met in Figs. 68-70.

**Merit Indices.** Similarly, various "merit indices" for calculating one over-all figure, purporting to express the grand total of the quality of a steel, appear farcical when only longitudinal values are considered.

One attempt to do this is by calculating a "P factor," by the

formula 
$$\frac{(6 \times \text{R.A.}) + \frac{\text{T.S.}}{1000}}{5}$$
 Two lots of steel of different degree

of directionality may show one to be much superior on the basis of the "P factors," calculated from longitudinal tests, whereas the "superior" one may be very deficient in R.A. in other than the longitudinal direction, hence decidedly the inferior one if R.A. means anything in service. Knowlton<sup>28</sup> states that, of all the merit indices, none is valuable. The conventional longitudinal tensile test puts the best foot forward.

**Size Effects.** Another way in which the best foot is put forward in many reports on properties conferred by alloying and by heat treatment, in quenched and tempered specimens, arises from the habit of reporting properties on material heat-treated in not much larger diameter than the standard 0.505-in.-diameter tensile specimen, and neglecting to impress the reader that a small specimen was used. Particularly in C steels, core hardenability decreases with extreme rapidity as the section increases, and so the structure and properties of the small specimen may be far superior to those of larger specimens.

To a smaller degree, there is an analogous size effect in as-rolled or normalized stock or in steel castings. A steel tensile specimen gives little useful data for prediction of the properties of larger pieces unless it has itself been taken from a known location in a large piece.

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## CHAPTER 6

### CONVENTIONAL TESTS—TENSILE, HARDNESS, ENDURANCE

Mechanical testing is done for one of two main purposes:

(a) For acceptance purposes to meet an imposed specification set up by someone who has decided that the requirements have meaning, or

(b) To try to evaluate, from the engineering point of view, the suitability of the material for a particular intended use, in which case the meaning and applicability of the tests are frequently in doubt, and remain in doubt through all tests, short of those of actual service or simulated service, in which service tests the "specimen" is in the exact shape, size, and heat treatment of the actual part.

Here the conventional tests may serve as road signs to indicate whether we are on the track trodden by previously used and satisfactory materials, though they may not prove that we have reached our destination.

In either case, attention must be given to testing methods to make sure that errors and differences not ascribable to the material itself have been eliminated. Then consideration must be given to natural scatter, so that enough tests are made to correct the impression that might be gained from an individual result on the high or low side of the scatter band.

With due attention to scatter, to sampling direction, and to size influences, the values from the conventional tensile test serve as a preliminary evaluation of engineering properties, though they leave unanswered many questions as to properties often more vital to service. They serve as a universal language for the first rough appraisal of the acceptability of a steel or of its heat treatment.

Quantitative tests for mechanical properties, commonly used as measures of quality, comprise tensile, hardness, fatigue, and notched-bar impact. The technique for making these tests has become quite standardized.

## THE TENSILE TEST—TENSILE STRENGTH

The tensile test determines the load the piece will carry under a steady pull before actually breaking. During the course of the test, there is also determined the load *below* which, if the load were released, the specimen would return elastically to practically its original dimensions and *above* which, on continued loading, it begins to yield and become permanently deformed, or, more commonly, the load at which a tiny, predetermined amount of permanent deformation occurs. After the bar is broken, the amount it has stretched in length and the amount it has contracted in cross section are measured.

These loads in pounds, divided by the original cross section of the bar in square inches, give the stresses in pounds per square inch corresponding to complete failure by fracture, that is, tensile strength, or ultimate tensile strength; and to elastic failure, that is, elastic limit, proportional limit, yield point, or yield strength, terms to be explained in the following paragraphs.

The elastic strength, naturally, is calculated on the original cross section of the bar. For convenience, the tensile strength is similarly calculated, but, as discussed on p. 128, this is a fictitious figure.

The true stress carried by the bar as the load increases from the yield stress to the ultimate while the bar is being pulled down to smaller cross section gives a useful picture of the behavior of the specimen during plastic deformation, but continuous measurement of the diameter during the test is somewhat inconvenient; hence, in the past, the true stress has seldom been determined. This situation is changing; true stress values are now quite frequently determined.

Abbreviations used in reporting tensile and yield strengths, to avoid repetition of "lb/in.<sup>2</sup>," are "psi," the Greek letter  $\psi$  (psi), or merely "tensile," with "lb/in.<sup>2</sup>" implied. The last-named, letting "lb/in.<sup>2</sup>" be implied, is used in this book.

The term "kip" is sometimes used to denote a unit of 1000 lb, thus 100,000 tensile might be reported as 100 kips/in.<sup>2</sup> The British report strength in long tons (2240 lb) per square inch and Continental observers in kilograms per square millimeter. The relation for conversion is  $1 \text{ kg/mm}^2 = 1422.32 \text{ lb/in.}^2$ , usually rounded off to 1420.

**Tensile Test Bars.** The standard round bar used for most heat-treated steels is, in the United States, 2 in. between gage marks and

0.505 in. diameter. This diameter corresponds to a cross section of 0.2 sq in., thus facilitating computation. The British bar is 0.564 in. diameter with 2 in. gage length, the German either 1 cm diameter and 10 cm gage length, or 2 cm  $\times$  20 cm. The French is 1.38 cm  $\times$  10 cm. The standard U. S. rectangular bar for soft structural-steel plate and the like is 8 in. between gage marks.

Figure 75 shows the standard 0.505-in. test specimen. Accepted dimensions for bars cut from pieces too small to give a standard bar

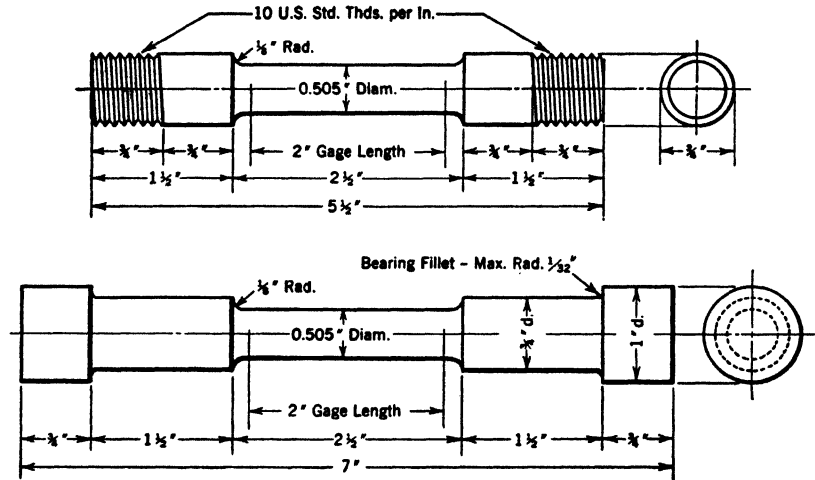


FIG. 75. Standard tensile test bars.

are shown in the *ASM Handbook*.<sup>1</sup> Suitable holders for bars without threaded ends may be used, provided proper precautions are taken for axial loading. Most tensile data in the literature are on specimens cut from 1-in. round bars. When heat-treated steels are tested at high hardness, the bar is usually machined to about  $\frac{1}{16}$  in. over the 0.505 diameter before treating and finally turned or ground to size. *If the steel does not harden throughout, the results on bars of different diameters will be widely different.* The structure of the test bar should be comparable to that of the section in which the material is to be used. In evaluation of steel to be used in very heavy section it is customary to test bars taken from the center, the midway point, and the edge, so as to survey the whole section. The usual specimen for cast steel is made in the form of a "keel" block with one or more projections about 1 in. wide, 1 to 2 in. high, and of length sufficient for one or two standard bars. Above the projections is a large mass of metal which is provided to ensure proper

feeding and avoidance of shrinkage. The keel block is shown in Fig. 76 and is fully described in the American Foundrymen's Association *Cast Metals Handbook*.<sup>2</sup>

It should be remembered that the 0.505 bar cut from a 1-in.-thick keel does not have the same rate of freezing as a heavier or a lighter section of an actual steel casting, and also that the thorough feeding of the test bar may not be paralleled in every part of the casting.

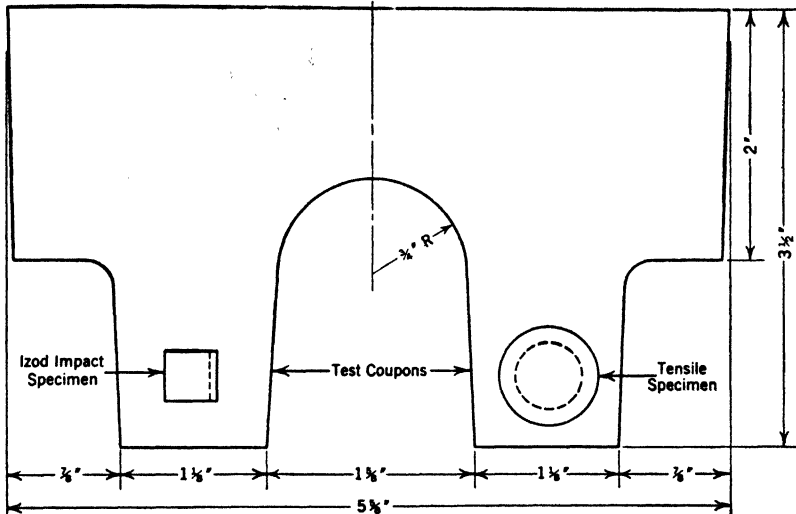


FIG. 76. Cross section of keel block for cast steel. The length may be as desired. A length of  $14\frac{1}{2}$  in. gives stock for two tensile test bars from a single keel. The feeding head may be made taller if desired; the height shown is the minimum.

Considerable care needs to be taken in interpreting the results of any tensile test into the strength of a structure using the steel evaluated by the test, since the specimen may not exactly represent the material in the structure. The purpose of the test is to establish the suitability of the material more often than the properties of the structure. This is especially true of cast steel where, because of difficulties of feeding, all parts of the casting will seldom be as sound and dense as the test bar.

**Yield Strength.** The method of determining the elastic limit or the yield strength needs to be clearly stated. The elastic limit is the stress at which the bar will just fail to show permanent elongation, that is, that value below which, if the load were released, the bar would return to its original length, thus showing all elastic and no plastic behavior. It is possible to determine this value by applying



and releasing progressively higher loads, remeasuring the length of the bar. When the stress above which the first perceptible permanent deformation occurs is so determined, the result is the "elastic limit," and the yield strength was formerly so denoted, though seldom actually determined in that way. As a "go" or "no-go" test, a "proof load" equal to the specified elastic limit is sometimes applied, no permanent stretch being permitted; but when no permanent stretch is found one does not know how much higher a load could have been endured.

Releasing the load slows down testing; but a determination can be made by noting the extension for each increment of load, and plotting strain against stress or getting the plot on an automatically recording testing machine. As long as the bar is behaving in a truly elastic fashion, stress is proportional to strain, and the plot will be a straight line whose slope is the modulus of elasticity. When plastic deformation starts, the line of the plot will begin to curve. The stress at the point of tangency of the curve and the straight line is the "proportional elastic limit." Unfortunately, the location of this point depends on the sensitivity of the measuring device used—the less sensitive it is, the higher the proportional limit appears to be.

Unless the sensitivity of measurement is very definitely specified and adhered to, different observers will report rather variant figures for proportional limit. The true proportional limit is lower than the practical load-carrying ability for most conditions of service in which a very tiny permanent deformation can be endured.

By slightly overstepping the yield strength in a soft ductile steel, the rate of elongation at a steady rate of loading is increased so fast that the beam of a lever-type testing machine will drop, the extensometer will show a marked increase in rate of motion, or a pair of dividers held in the gage marks will tend to jump out. This crude "yield point by drop of beam" has been much used in rapid rough-and-ready testing but is not applicable to many of the strong heat-treated steels whose stress-strain curves merge gradually from the straight line into a curve without the very sudden extension of the softer steels. Figure 77 shows typical stress-strain diagrams for a soft and a hardened steel.

A method is needed which will compare the elastic properties of all steels, both the very soft and the very hard. This is attained by setting up a definite amount of permanent elongation as the criterion of yield strength. This is generally either 0.1 or 0.2% of the 2-in.

gage length. The report should state which one is used; 0.2% is more common and is implied when not stated. This specified set is laid off on a plot of the stress-strain diagram as an intercept, and a line with that intercept is drawn parallel to the beginning of the stress-strain line for the steel. This line cuts the stress-strain curve at the yield strength. Different observers can readily get duplicate values for yield determined in this way; so the method is coming into general use and the term yield strength is supplanting the older

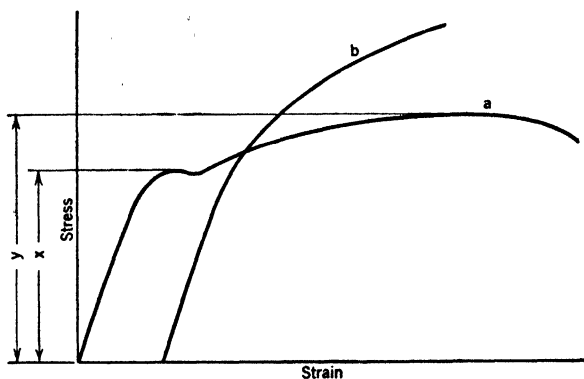


FIG. 77. Typical stress-strain diagrams.

- a—for soft steel
- b—for hardened steel
- z = yield strength for curve a
- y = tensile strength for curve a
- $\frac{z}{y}$  = yield ratio for curve a

terms elastic limit, yield point, and proportional limit. Figure 78 illustrates the application of this method.

Another way of setting a value for that amount of permanent deformation to be considered as the yield strength is to draw a tangent to the curve, the tangent having twice the slope of the elastic modulus line. The point of tangency is the "Johnson" elastic limit, as shown in Fig. 78. Depending on the shape of the stress-strain diagram, the Johnson limit may be either above or below the value given by the intercept method.

**Breaking Strength-Necking.** After the yield strength is passed, loading is continued, the bar stretches and, except in very hard steels, finally necks down and, of course, finally breaks. The maximum load, point a in Fig. 77, divided by the original cross section, is the tensile strength. The load at fracture, as Fig. 77 shows, may be lower than this maximum load. Dividing the load at fracture

by the final cross section measured after fracture gives the "breaking strength," the maximum strength after extreme cold work by stretching, but this figure is seldom included in the usual test of tensile properties. In fact, this "breaking strength" is a fictitious value, since, once a neck has formed, the stress is no longer uniform, nor is it a simple stress.

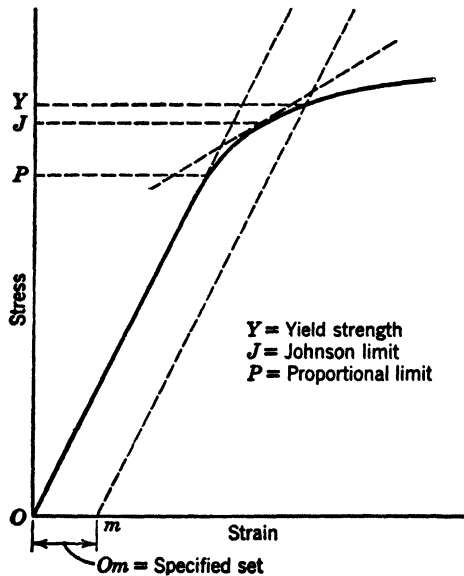


Fig. 78. Intercept method of determining yield strength (after ASTM) and Johnson limit.

**Work Hardening during the Test.** The course of the curve from the yield stress to the rupture stress is affected by the way the material work-hardens, and this has a lot to do with cold formability. Since the piece is stretching and becoming smaller in cross section as extension goes on, and the usual plot takes no account of this, that plot gives no inkling of these changes.

A comparison of the final measurement of elongation and of contraction gives a hint as to what has happened. If the material work-hardens readily, its elongation is spread quite uniformly over the test length, and it does not neck down locally at the fracture to an extreme degree. If it does not work-harden readily, the piece necks sharply, and the elongation is much greater close to the fracture than at a distance from it. Figure 79 shows the general extension of the readily work-hardened 18:8 stainless steel.

More insight is gained when special optical or other devices are used to follow and record the minimum cross section at the neck, so

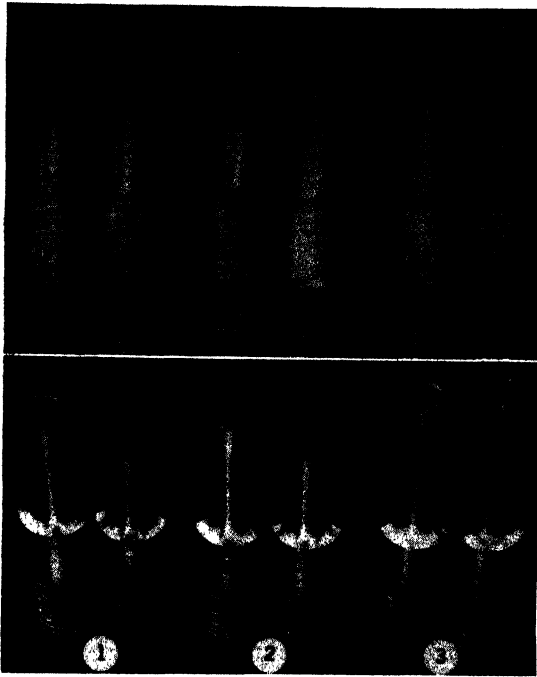


Fig. 79. Tensile bars of three different types of material after fracture: Normalized wrought ferritic steel, quenched and tempered wrought ferritic steel, and wrought austenitic steel. Bars 1 and 2 have the following composition:

C	Mn	Si	Cr	Ni	Mo	Al *	B *
0.30	0.82	0.21	0.51	0.52	0.31	0.10	0.003%

\* Amount added.

Bar 3 is a section of a commercial bar of 18-8 stainless steel.

Bar 1 was normalized 1 hr at 1700°F.

Bar 2 was normalized 1 hr at 1700°F, reheated to 1600°F, held 1 hr and water-quenched, followed by tempering 2 hr at 1000°F.

Bar 3 was not heat-treated.

*Mechanical Properties*

	Bar No.		
	1	2	3
Tensile strength	126,000	162,250	96,750
Yield strength	88,000	148,500	42,500
Reduction area, %	50.6	54.9	66.3
Elongation, %	18.5	17.0	65.5
Hardness, R.C.	23	33	.....

The quenched and tempered bar has greater reduction of area than the normalized bar, in spite of the fact that it is 10 points R.C. harder.

that the stress endured at any moment can be plotted against the minimum section existing at that moment, in a "true stress-strain curve." Also, by marking off a specimen into several gage lengths

and recording the final elongation and reduction of area over each part of the gage length, a better picture of deformability is secured. This, however, comes into the category of special tests, rather than conventional ones.

**Ductility.** The deformation produced in fracturing the bar is, of course, a measure of its ability to be deformed, that is, its ductility. Measurements are made of the change in length and the change in cross section.

Ductility is a rather general term. High ductility usually connotes formability and toughness, but other more direct methods such as bend and impact tests are used to supplement the ductility figures when these properties are to be evaluated.

Elongation and reduction in area (contraction) are reported in percentages of the original dimensions; and will be herein abbreviated to % elong. and % R.A. The % elong. depends on the dimensions of the bar as well as on the properties of the steel.

**Effect of Dimensions.** Tensile and yield strengths determined on the different standard bars are closely intercomparable.

It may be noted, however, that the elongation measurement includes the portions near the neck. Where deformation has not been uniform, this makes the elongation dependent on the ratio of length to diameter.

The elongation of the same steel tested in different dimensions, such as those in vogue in different countries, is not exactly the same, being affected by the ratio of length to diameter. Janitsky<sup>3</sup> has suggested methods for calculating what the elongation would be on a bar of one size when another size is used. More data than are usually reported are necessary to make a useful conversion when steels are compared whose tendencies toward general elongation and local necking are widely different. Reduction of area, at least when the necking down occurs near the middle of the bar, is not affected by the length; so this percentage figure is comparable for all round bars. Elongation values given herein are on the 2-in. gage length unless otherwise stated.

Strength and ductility are inversely related: the higher the strength the lower the ductility. The more ductile the steel is, the weaker it will be. The exact figures differ with the structure, hence with the heat treatment, but within the same class of structure, the ductility is roughly predictable from the strength, and vice versa. The engineering problem is to select that particular compromise best serving the use. Some uses demand strength and little or no

deformability: others call for extreme deformability, strength being a secondary matter; still others fall in between.

**Yield Ratio.** The yield strength is not predictable from the tensile-strength figure by itself, for the yield ratio (see Fig. 77) shifts with the composition, heat treatment, and structure of the steel. But, again, within the same class of steels, whose range of yield ratios is known, the tensile-strength value does give a strong clue to the yield strength. The best yield ratio comes from quenching to 100% martensite, and then tempering to required softness. A smaller proportion of martensite in a slack-quenched piece gives lower yield strength at a given tempered hardness.

Thus, if we are acquainted with the class and structure of the steel, and know how it has been heat-treated, knowledge of the tensile strength carries with it an implication as to the yield strength, as well as the ductility and toughness normally associated with that strength (see Figs. 68-70).

**General Precautions in Tensile Testing.** For accuracy and reproducibility in tensile testing, certain precautions are necessary. The testing machine must be properly calibrated and properly sensitive to changes in load. The specimen must be of standard dimensions (or, if a standard specimen cannot be taken from the material to be tested, any deviation must be reported), the specimen must be gripped so that it is loaded with reasonable axiality, and the rate of application of the load must be held within limits.

The requirements as to testing machines and their calibration and those on methods of tension testing are authoritatively set forth by the American Society for Testing Materials.<sup>4</sup>

Axiality of loading, in a machine that is itself properly aligned, is usually obtained by the use of threaded grips. Specimens too hard to be threaded can be made with shouldered ends and held in fixtures with spherical seats or ball-bearing alignment devices. Wedge grips, which serve on very soft and ductile materials, are not suitable for most heat-treated products.

The speed of pulling of the specimen, up to the yield stress, must not exceed  $\frac{1}{8}$  in. per minute. Faster pulling leads to erroneously high results. Even the  $\frac{1}{8}$  in. per minute is far too fast for some special alloys. The testing temperature is not ordinarily specified, since it is expected to lie within the normal room-temperature range, and the effect of such variations is ordinarily not great. It cannot be assumed without proof that this is the case for all steels,

and some observers think that R.A. values tend to be low in cold weather.

**Modulus of Elasticity.** In the plot of a tensile stress-strain diagram, the modulus of elasticity is shown by the slope of the line below the elastic limit. The modulus enters into many engineering calculations,<sup>81</sup> which generally take it as 29,000,000 psi.

Few steels have moduli that deviate much from this value, and the modulus cannot be materially altered by heat treatment. When it becomes necessary to determine the modulus with exactness, test bars of long gage lengths are used and precision extensometers employed. Routine tensile tests on standard specimens do not give sufficiently accurate data for such a determination.

### HARDNESS TESTING

Hardness is resistance to penetration. It is most often commercially measured by pressing a harder object into the metal to be tested and noting the penetration. For soft steels the penetrators are usually hardened steel balls; for harder ones, tungsten carbide balls<sup>5</sup> or, more commonly, diamond points, cones, pyramids, or rhombs.

It happens that tensile strength can be quite closely predicted from hardness, and hardness is very easy to measure. A hardness determination can often be made on a finished part without injuring the part for service. In such cases it is a nondestructive test.

The quench hardenability of steel is also roughly indicated by the hardness-distance curve of the Jominy bar, though the hardness is not a sure indicator of the structure. The effectiveness of a quenching medium is indicated by the depth of hardness produced in a quenched specimen, and the response to tempering is likewise indicated by hardness measurements on tempered pieces. Within like steels, in known conditions of heat treatment, the hardness reflects the carbon content.

Thus, if knowledge of the steel's expected structure and of the properties accompanying it is presupposed, a hardness determination can give much information and will often indicate whether there has been a mix-up in steels or a slip in heat treating.

**Brinell Hardness.** The Brinell hardness number (abbreviated to BHN, or simply Brinell) is calculated from the diameter of the impression left after a 10-mm diameter ball (supposed to be rigid and not itself to deform) is pressed into the specimen vertically in

not less than 6 sec and then held at the maximum load of 3000 kg for 30 sec. The Brinell number is calculated as the load divided by the spherical area of the impression. The harder the steel, the smaller the indentation for a given load, and the higher the Brinell. For many purposes the indentation diameter can be used, without transferring this into the Brinell number. Table 5 gives Brinell numbers calculated from the diameters. The diameters, the load,

TABLE 5

## BRINELL HARDNESS NUMBERS

Diameter of Steel Ball = 10 mm

Diameter of Ball Impression, Mm	Hardness Number for a Load of 3000 Kg	Diameter of Ball Impression, Mm	Hardness Number for a Load of 3000 Kg	Diameter of Ball Impression, Mm	Hardness Number for a Load of 3000 Kg	Diameter of Ball Impression, Mm	Hardness Number for a Load of 3000 Kg	Diameter of Ball Impression, Mm	Hardness Number for a Load of 3000 Kg
2	946	3	418	4	228	5	143	6	95
2.05	898	3.05	402	4.05	223	5.05	140	6.05	94
2.10	857	3.10	387	4.10	217	5.10	137	6.10	92
2.15	817	3.15	375	4.15	212	5.15	134	6.15	90
2.20	782	3.20	364	4.20	207	5.20	131	6.20	89
2.25	744	3.25	351	4.25	202	5.25	128	6.25	87
2.30	713	3.30	340	4.30	196	5.30	126	6.30	86
2.35	683	3.35	332	4.35	192	5.35	124	6.35	84
2.40	652	3.40	321	4.40	187	5.40	121	6.40	82
2.45	627	3.45	311	4.45	183	5.45	118	6.45	81
2.50	600	3.50	302	4.50	179	5.50	116	6.50	80
2.55	578	3.55	293	4.55	174	5.55	114	6.55	79
2.60	555	3.60	286	4.60	170	5.60	112	6.60	77
2.65	532	3.65	277	4.65	166	5.65	109	6.65	76
2.70	512	3.70	269	4.70	163	5.70	107	6.70	74
2.75	495	3.75	262	4.75	159	5.75	105	6.75	73
2.80	477	3.80	255	4.80	156	5.80	103	6.80	71
2.85	460	3.85	248	4.85	153	5.85	101	6.85	70
2.90	444	3.90	241	4.90	149	5.90	99	6.90	69
2.95	430	3.95	235	4.95	146	5.95	97	6.95	68

and the time are all capable of accurate measurement; that is, the method is not dependent on anything that cannot be quite readily checked.

For steels of 70 to 500 Brinell, hardened, accurately ground, and polished 10-mm balls of steel of the regular ball-bearing type are commonly used. Use on harder steels permanently deforms the ball



itself; the first impression made with the ball gives a hardness lower than the true one, and in subsequent use the results become even more inaccurate. Work-hardened balls, which decrease this error, are obtainable, with which little inaccuracy results up to 600 Brinell, and carbide balls can be had that will serve to around 800.

For very soft materials, a 500-kg load is used. Brinell numbers can be computed by the formula

$$BHN = \frac{P}{\frac{\pi D}{2} (D - \sqrt{D^2 - d^2})}$$

where  $P$  is pressure in kilograms, and  $D$  and  $d$ , respectively, the diameters of the ball and of the impression in millimeters.

So-called "Baby Brinell" tests with tiny balls are sometimes made on objects too small for the standard test. The data computed from nonstandard conditions usually differ somewhat from those obtained under standard conditions, and, when any load, ball, diameter, or ball material other than the usual ones is employed, the fact should be stated.

Precautions<sup>6</sup> necessary in Brinell testing are to see that the loading device shall press the ball upon the specimen in a truly vertical direction, to observe the required rate and time of loading, and to make sure that the maximum load is reached but not exceeded. That is, the Brinell machine must be kept in calibration within an error of  $\pm 10$  kg in 3000. The ball must have the nominal diameter (10 mm) within 0.025 mm (0.001 in.) and be discarded if deformed beyond that limit. The measuring microscope must read true dimensions to 0.01 mm.

The diameter of the impression must be determinable within an error of 0.01 mm. This requires that the specimen be flat, smooth, and freed from deep tool marks; therefore smooth machining, filing, or grinding is often necessary. If the specimen is curved, the radius of curvature should be at least 25 mm (1 in.), and the diameters of the impression must be averaged, since it is not quite circular. The center of the impression must be at least three of its own diameters away from the edge of the specimen or the center of another impression. The specimen, for test under the 10-mm ball and 3000-kg load, must be at least 0.4 in. thick. These precautions are more fully described in several accessible sources.<sup>7-9</sup>

**Monotron.** Another ball-indentation tester, the Monotron, uses a tiny diamond hemisphere. Instead of applying a definite load and

reading the diameter of the impression, increasing load is applied until the impression has penetrated to a standard depth.

The scale is usually calibrated in terms of the equivalent Brinell numbers. Because of the light load, shallow impression, and use of the diamond ball, thinner and harder materials may be tested than can be handled by the regular Brinell test. The results at high hardness would be more reliable than those obtained with a deformable ball.

**Work Hardening, the Meyer Analysis.** Hardness tests can be so carried out as to evaluate the ability of the metal to be work-hardened. The most common method of using hardness tests to provide information on work hardening is the so-called "Meyer analysis," after an early experimenter, who studied it.

To apply this method, a series of three or four Brinell impressions is made. The first impression is made with a load much lighter than the standard 3000-kg load (say 500 kg). Succeeding impressions are made at higher loads up to the 3000-kg standard. The diameter of each impression is measured and, when the load in kilograms is plotted against the diameter of the impression in millimeters on log-log paper, a straight line results, which may be fitted to the equation,  $\text{load} = ad^n$ , where  $d$  is the diameter of the impression in millimeters and  $a$  and  $n$  are the "Meyer" constants. The constant  $a$  is determined directly from the curve and is the load for which the impression diameter is 1 mm. This constant (divided by 0.785 to put it in kilograms per square millimeter) has the dimensions of a "Brinell" number and may be interpreted as the initial hardness before appreciable work hardening has occurred. It will be a much lower figure than the regular Brinell figure, at the normal 3000-kg load.

The slope of the log-log plot, the exponent  $n$  of the equation, is the measure of the work hardenability of the metal. When a metal is completely work-hardened, that is, "full hard," the exponent is 2. That for an annealed metal with high work-hardening capacity may be 2.5 or even higher.

Where the Meyer analysis is performed as described, it will sometimes be found that the logarithmic relation between load and impression diameter is not linear. This is usually caused by local inhomogeneities, and to avoid this discrepancy some investigators prefer to make the successive impressions in the same spot. This is difficult to accomplish, however, unless it is possible to measure the diameter of each impression without moving the test piece. The

Meyer constant  $a$  and the Meyer exponent  $n$  serve to define the hardness and work-hardening properties more fully than does the regular Brinell value taken at one load. Technical literature often uses the Meyer values in discussions of work hardening. Figure 80, after Heyer<sup>10</sup> and O'Neill,<sup>7</sup> shows the way  $a$  and  $n$  are related to each other and to the tensile strength.

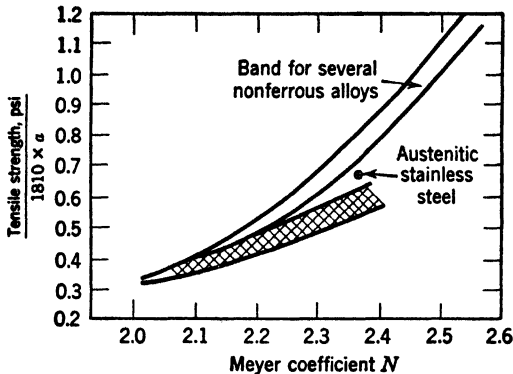


FIG. 80. Hatched band for steel (data from Heyer and O'Neill). Relation between strength  $\div$  Meyer constant  $a$ , and Meyer exponent  $n$ .

**“Cloudburst” Hardness Testing.** To reveal soft spots on a hard surface, a shower of steel balls may be dropped on the specimen or, in production control, on an actual part. By regulating the height of fall, the balls can be made to mark the soft spots without appreciably indenting the hard ones. Such a test may be looked on as a special case of shot blasting. Details are discussed in the *Metals Handbook*.<sup>1</sup>

Actual shot blasting is advocated by Peterson<sup>11</sup> for inspection of high-speed steel tools and coil-spring material.

**Mutual Indentation Hardness Tests.** Instead of using a hard ball or other penetrator, hardness measurements can be made by pressing two suitable specimens of the same material together. The data can be calculated into Brinell values. Crossed prisms or cylinders are sometimes used; but two cylinders, lying parallel and in contact throughout their length, one above the other, held in suitable blocks to maintain alignment, are more commonly used. The pressure applied flattens the portions of the two cylinders in contact. Measurement of the flat areas, together with the pressure applied, give data from which Brinell hardness may be computed. Cowdrey<sup>12</sup> has described the method in detail.

**Hot Hardness.** The mutual indentation method is seldom applied in ordinary hardness testing. It finds chief application to determinations at elevated temperatures. Harder and Grove<sup>13, 14</sup> developed and described the method for high temperatures. It has been successfully used in the study of high-speed steels, valve and die steels, and other materials for high-temperature service.

An alternative method is to use the regular Brinell test with a ball that does not deform appreciably at the test temperature. Tungsten carbide balls have been so used by several workers. Fetzer<sup>15</sup> has described equipment and methods for high-temperature hardness testing by impact of a falling weight carrying an indenter, for study of behavior without the flow produced by holding the load in a static test. Bishop and Cohen<sup>79</sup> discuss Rockwell cone tests at elevated temperatures.

Bens<sup>90</sup> describes an elaborate equipment for making Vickers hardness tests up to 1700°, in a vacuum, and presents evidence that hot hardness is a useful index of the high-temperature strength of stable alloys. Grant,<sup>91</sup> on the other hand, dismisses hot hardness as "of little value as a measure of high-temperature-alloy suitability," but records no data by which the reader may judge the accuracy of his appraisal.

**Vickers Pyramid Test.** An indentation hardness test applicable over a wider range of hardnesses than the ball hardness method is the diamond pyramid test, used in the Vickers and some other instruments, results of which are most commonly reported as Vickers or Vickers-Brinell numbers (VPH, Vickers pyramid hardness, or DPH, diamond-pyramid hardness). The point of a small diamond pyramid (136°) with a square base is impressed into the specimen just as is done in Brinell testing, but under a very much lighter load. The impression is much smaller than in the Brinell test; hence, much thinner specimens may be tested. The impression tends to be more sharply outlined and its diagonal more easily measured than a spherical one, though on account of its small size a higher magnification is required. The hardness number is calculated, as in the Brinell test, from the load and area. The diamond point does not deform plastically and deforms only slightly elastically, so that the calculated hardness is not appreciably affected by use of different loads; hence the depth of indentation can be regulated to suit the thickness of the specimen. It need be only of the order of 0.002 in. The means of applying the load can, of course, be any suitable device, whether the Vickers machine itself or not.

The formula for calculation of Vickers hardness is

$$\text{Hardness} = 1.8544 \frac{\text{load}}{(\text{diag.})^2},$$

the load being measured in kilograms and the diagonal of the impression in millimeters.

The advantages of the Brinell and Vickers-Brinell methods, in that all factors can be measured in absolute units, are somewhat lessened by the necessity of measuring the impression, which slows up the operation.

Devices are now on the market by which the image of the Brinell or Vickers impression is projected on a ground-glass screen, against a scale, for quick reading.

An advantage of the pyramid method is that, by varying the loads, hardness determinations may be made on the softest and the hardest alloys, and the results are expressed on a single scale without the necessity for shifting from one scale to another as in most of the other methods.

**Firth.** The Firth hardnessmeter is similar to the Vickers, using a diamond pyramid and measuring the diagonal of the indentation. The figures obtained are the same as with the Vickers.

**Eberbach.** The Eberbach instrument uses a spring-loaded Vickers pyramid or Knoop rhomb mounted in a tube, interchangeable with the barrel of a microscope. Down to 300 VPH Eberbach results agree well with Vickers; in softer materials, the higher the load on the penetrator, the higher the apparent hardness. Bernhardt<sup>16</sup> has discussed the details of its use, and Robinson<sup>17</sup> gives a nomogram for reading the hardness from the measurement of the diagonal of the pyramidal impression.

**Knoop-Tukon Test.** Knoop, Peters, and Emerson<sup>18</sup> devised a diamond indenter of the shape shown in Fig. 81. This Knoop indenter produces a shallow indentation with the impression of the long diagonal and is about three times as long as one made by the Vickers pyramid at the same load. The Knoop diamond is generally mounted in the Tukon tester, which applies a known weight through a balance beam. The load is applied for 20 sec. The diagonals of the resulting indentation are measured with a micrometer microscope. The long diagonal does not change dimension when the load is removed, but the short diagonal becomes shorter because of the elastic recovery of the material. Thus the "unrecovered" and "recovered" areas in the plane of the surface can be

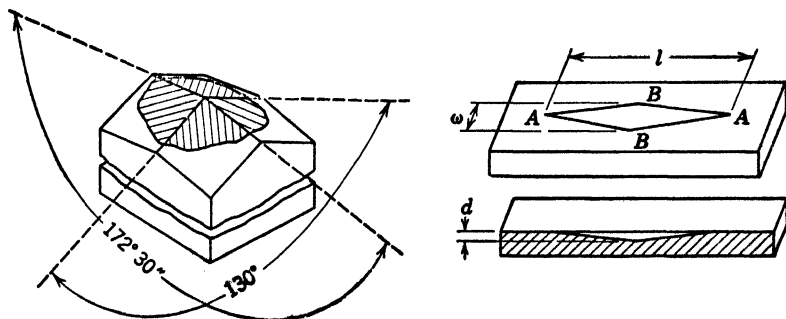


FIG. 81. The design of the Knoop indenter and shape of the indentation.

calculated. The Knoop hardness (KHN) is obtained by dividing the load in kilograms by the *unrecovered* area as follows:

$$I = \frac{L}{Ap} = \frac{L}{l^2 Cp}$$

$I$  = Knoop hardness number

$L$  = load (in kilograms) applied to indenter

$Ap$  = unrecovered projected area of indentation (in square millimeters)

$l$  = measured length of long diagonal of the indentation (in millimeters)

$Cp$  = constant relating  $l$  to the projected area. For an indenter of  $172^\circ 30'$  longitudinal angle and  $130^\circ 0'$  transverse angle,  $Cp$  equals  $7.028 \times 10^{-2}$

A table is provided by the makers showing the hardness corresponding to various lengths of the impression.

The load is varied according to the hardness of the material to be tested. With loads of 200 g or over, the Knoop hardness figures do not vary much with the load and coincide with those of the Vickers in the case of the usual heat-treated steels.

At very low loads, the Knoop values depend on the load, being the higher at the lower loads, so that the load used should be stated when the values are reported.<sup>19, 20</sup>

An exhaustive study of Knoop testing, especially on superhard nonmetallic materials, but including hardened tool steel, is described by Thibault and coworkers.<sup>22, 23</sup> Shubrooks<sup>24</sup> discusses Knoop testing of very tiny objects and gives a Rockwell-Knoop conversion chart.

Figure 82 shows the relative size of the Knoop and Vickers impressions in an austenitic valve steel. at 100- and 500-g loads.

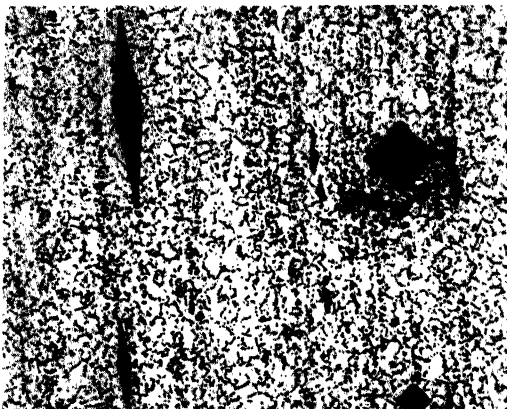


FIG. 82. Knoop and Vickers impressions in an austenitic valve steel.  $\times 200$ .  
(Battelle)

**Rockwell-Cone Test.** To hasten the testing, a method somewhat similar to the Vickers has been devised in which a diamond cone ( $120^\circ$  with a slightly rounded tip), instead of a pyramid, is used as penetrator; but the depth of penetration is measured by the testing instrument, instead of one's waiting to measure the breadth of the impression. This is the Rockwell-cone hardness tester. In the Rockwell test the penetrator is first seated by a small (so-called "minor") load, the dial set to zero, the large (major) load (regularly 150 kg) applied for a specified time, the major load removed, and the hardness number is then read directly from the dial.

Each instrument has to be separately calibrated in arbitrary units, and any difference in the shape of the penetrator tip will affect the calibration. This type cannot be so certainly checked in absolute units as can the Brinell or diamond-pyramid types but has the advantage of great rapidity, a reading being made in a few seconds.

The depth of penetration is of the order of 0.005 in., so that thin specimens may be tested. In order to make it feasible to test still thinner hard surfaces, like those on nitrided steel, a similar piece of apparatus, the Rockwell superficial-hardness tester, has been devised in which the depth of penetration is only about 0.002 in. and the depth-measuring device is correspondingly more sensitive. Figure 83 shows the relative depths of penetration.

The Rockwell testers can be used with different minor and major loads, and these have been varied for work on material of different hardness ranges and scales, differently lettered, provided for such cases. Hence, the scale used must always be stated. Unless a different scale is stated, the "C" scale, the regular penetrator, and the regular major load of 150 kg are implied. An abbreviation often used is Rc.

Besides the cone penetrator, the Rockwell machine is provided with a  $\frac{1}{16}$ -in. diameter steel ball, used with a major load of 100 kg. This gives a type of "Baby Brinell" test more rapidly carried out, because measurement of depth is done by the machine itself. The

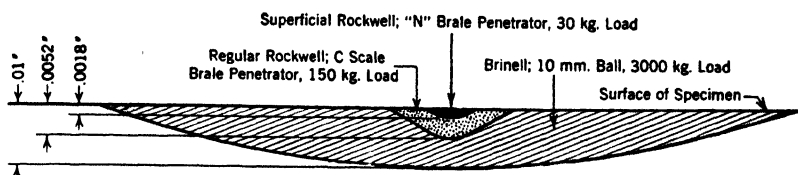


Fig. 83. Relative depths of Brinell, Rockwell cone, and Rockwell superficial hardness impressions.

Rockwell "B" test is only employed on soft steels, not harder than 300 Brinell. Detailed descriptions of the Rockwell method are available.<sup>1, 21</sup>

Rockwell determinations on curved surfaces will be low. Wallace<sup>22</sup> gives plots for the correction to be added to the ball, C, 30N and Superficial 30T readings for cylindrical specimens of various diameters. Similar data are given by Fleischmann and Jenkins.<sup>23</sup>

**Shore Rebound Test.** Another type of diamond-pointed indenter and a different method of measurement are used in the Shore Scleroscope. A diamond-tipped hammer is dropped from a definite height upon the surface to be tested, and the hardness is measured by the height of the rebound, the scale being graduated in arbitrary units. The depth of penetration is very small, so that the instrument can be used on case-hardened specimens. The rebound is regulated more by the elastic limit of the specimen than by its tensile strength, so that the Shore does not measure exactly the same type of hardness as do the indentation methods. The Scleroscope requires a very smooth surface on the specimen, which must be solidly held, in a horizontal position, and the hammer must fall exactly vertical. Different instruments tend to give slightly different readings, and it is important to check the readings often upon standard blocks. Several Shore read-



ings should be taken and averaged, care being taken not to let the hammer fall twice in the same place.

The Shore instrument was once widely used upon steels too hard for good Brinell readings, but with the advent of the diamond cone and pyramid penetrators its use has been less common. It is especially useful for testing large chilled rolls, too large to take conveniently to the other types of equipment, for the Shore instrument can be transported to the work.

Kommel<sup>23</sup> has emphasized the care necessary in calibration and use of the Scleroscope in order to obtain useful results.

**General Precautions in Hardness Testing.** Certain precautions are necessary with any of these hardness tests. The load must be applied truly perpendicular to the specimen to be tested. The surface should be flat and all scale or decarburized surface removed, the latter precaution being the more important the shallower the indentation. The surface resting on the anvil should make good contact so as to avoid springiness in thin specimens. Curved surfaces of fairly large radius of curvature in comparison with the width of the indentation may be tested, but results will not be exactly the same as on a flat surface. With the Brinell test, the long and short diameters of the impression on a curved surface are measured and averaged. The smaller the indenter, the more carefully smoothed and polished must the specimen be, the 10-mm Brinell ball being the least rigorous in its requirements, the Scleroscope the most.

No diamond point can be expected to give true results if it is chipped, and precautions as to vertical loading, smoothness of applying the load, etc., are needed for the sake of the diamonds as well as for accuracy. All hardness testers should be periodically checked against standard blocks of known hardness and uniformity and the instruments kept in calibration. Those that have empirical scales, such as the Rockwell and Scleroscope, should be returned to the makers when out of adjustment, since they are not readily calibrated by the user.

In all hardness testing the specimen must be firmly supported. This, of course, is especially vital for thin specimens. Very great attention has to be given to the "anvil effect" in testing very thin specimens, and there is much discussion in the literature as to the limiting thickness for accuracy in the various types of test, especially upon soft sheets. In heat treatment the objects are usually massive enough so that no trouble is encountered, though shallow carburized

or nitrided cases involve analogous problems. Williams<sup>24</sup> discusses these problems.

Malcolm<sup>25</sup> gives an illuminating set of curves showing errors in attempting to measure the hardness of thin cases with indenters under too heavy loads. For example Rockwell C readings at the surface and 0.01 in. below the surface on a particular thin case were 54 and 33. The corresponding Rockwell 15 N readings were 88 and 85. Similar results are given by Beckwith.<sup>26</sup>

In careful work with any of these types of hardness testers, several readings, usually four, are made and averaged.

**File Hardness.** Besides the quantitative hardness tests, a much-used and useful shop test is filing with standard testing files. The personal equation of the man handling the file, the angle at which it is held, the force applied, etc., are variables; but, by frequently checking himself by filing standard blocks and discarding test files as soon as they become worn, the tester can make accurate comparisons of the hardness of steels, from, say, 60 Rockwell C up. The usual testing files will not "touch" quenched steel of 64–68 Rockwell C but will bite quenched steel that has been tempered at 375° to 64–66 Rockwell C. It is possible to make files that will cut hardened steel tempered at 400° and not that tempered at 375°, and still further gradations are possible.<sup>1, 27</sup>

**Scratch Tests.** A type of file testing may be carried out by scratching with a sharp needle point, the point usually being ground on the tip of a square tapered file made from plain C steel with a hardness, as quenched, of 65 Rockwell C. By tempering at a series of temperatures, needles of a graded range of hardness can be obtained. This method is described by Hamilton<sup>28</sup> for use on materials or surfaces so thin that other methods are not applicable. More elaborate scratch tests are made by dragging a diamond or sapphire point under a definite load, over a surface, polished as for metallographic examination, then measuring the width of the scratch. A device for this test is sold under the name "Microcharacter." The method is used in research work to differentiate between microconstituents of an alloy. It is suggested by Conley and coworkers<sup>29</sup> as of possible applicability to the study of carburized or nitrided cases and, since the test is itself one of cutting, to evaluation of machinability. The method is seldom applied for commercial evaluation of hardness.

**Pendulum Hardness.** Another type of hardness test equipment, devised by Herbert, consists of an apparatus with a hard ball near the center of gravity of a heavy framework. The whole apparatus is

rested on the specimen through the ball, thus indenting the specimen. When the apparatus is tilted and released, it swings like a pendulum. The angle of swing after release and the time for a definite number of oscillations are alternatively taken as measures of hardness. Claims have been made that these values give information not brought out by other testing methods. Much care and skill are required to get concordant check readings. This device has found practically no commercial use in the United States.

Many other types of hardness testers have been devised for routine or research work, some of which are as useful as those previously mentioned, others useful only in special cases.

**Hardness Conversions.** It would be highly desirable to be able to convert accurately hardness figures obtained by one method into those obtained by another. Many investigators have made such comparisons on the machines available to them and have given conversion formulae, tables, or curves. Any one investigator is likely to get quite smooth curves, applicable to his own machines, as long as they are in calibration and as long as the types of steels compared do not vary too widely. The results of different investigators are prone to deviate considerably, especially in the higher hardness range. When such different results are plotted, the curve becomes a band rather than a line, the width of the band representing the degree of uncertainty likely to be encountered among commercial machines in good repair. Nevertheless, the average curve obtained from all these data affords a convenient conversion, usable for general purposes.

The difficulty of making exact conversions is shown by the shaded bands in Figs. 84–88. Figure 88 includes an area *ABCD* for the scatter found by Hengemühle and Clauss<sup>30</sup> for seven German Shore-type instruments. Shore data given by Kommel<sup>28</sup> (Fig. 89) fall within the shaded band of Fig. 88, but deviate somewhat from the heavy line in that figure. He emphasizes the precautions necessary for obtaining concordant readings from different Shore instruments.

It is obvious from Fig. 86 that ordinary Brinell values are increasingly in doubt above 400 Brinell because of elastic and plastic deformation of the ball itself. Balls of higher elastic properties behave fairly well up to 500 Brinell. Higher hardnesses are best determined with diamond balls, as in the Monotron, or, more commonly, by diamond points.

Within the range 300 to 600 Brinell, Rockwell cone values are approximately  $\frac{1}{10}$  of the Brinell value.

It is now customary to determine and report hardnesses above 400 Brinell in Vickers, Knoop, or Rockwell C values. According to Martin and Wiley<sup>31</sup> and Lysaght,<sup>20</sup> Knoop values, using the 500-g load,

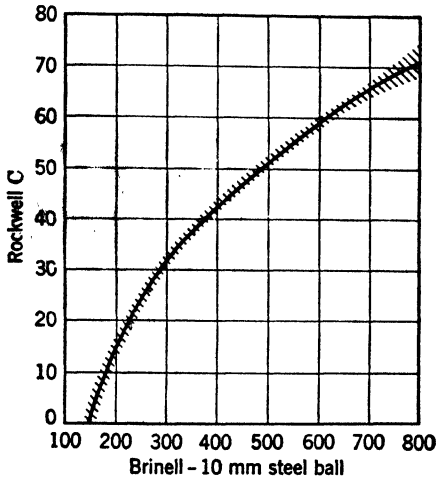


FIG. 84. Conversion chart. (Brinell and Rockwell C)

are equal to Vickers numbers and fall within the scatter band of Fig. 87.

As Fig. 87 shows, hardness determinations on superhard materials tend to be erratic and difficult to duplicate and convert from one

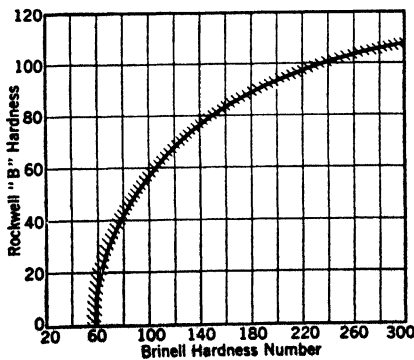


FIG. 85. Conversion chart. (Brinell and Rockwell B)

system to another. According to different investigators, 68  $R_c$  may mean anything from 875 to 1200 Vickers. In this hardness range, the Knoop test is increasingly used. Many conversion charts and

tables have been published by Hruska,<sup>82</sup> Vanick and Eash,<sup>83</sup> Clapp and Clark,<sup>84</sup> Scott and Gray,<sup>85</sup> Gill,<sup>86</sup> *The Metals Handbook*,<sup>1</sup> etc. In some of these the various Rockwell scales (see also Fig. 92) and the Monotron values are included.

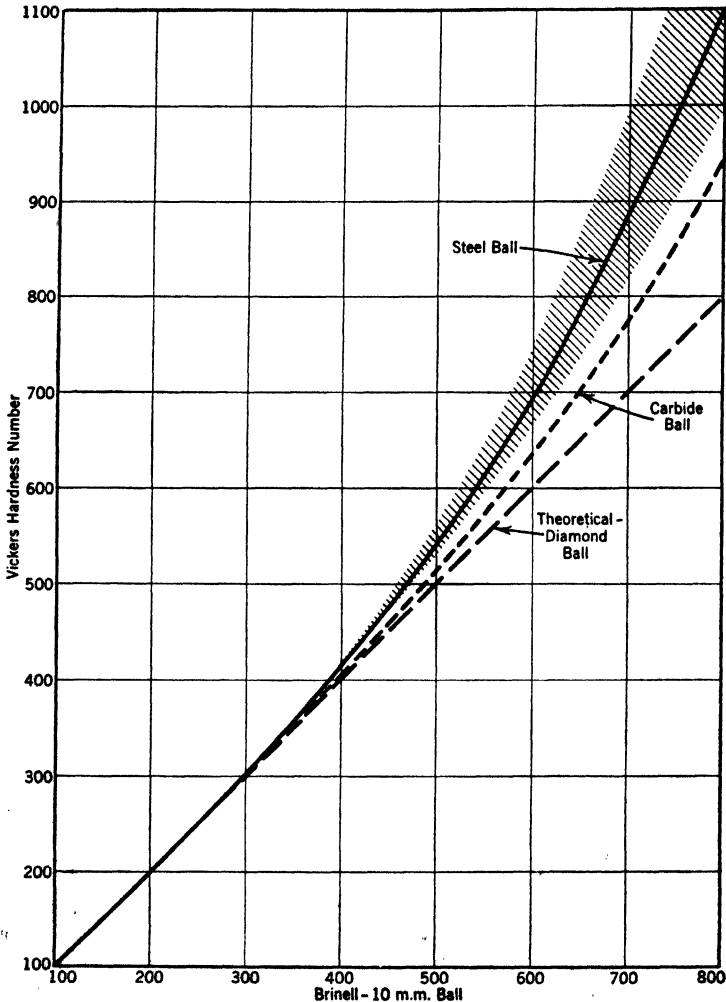


FIG. 86. Conversion chart. (Brinell and Vickers)

Gill<sup>86</sup> points out that conversions worked out for ordinary constructional steels are inaccurate for tool steels and that C tool steel and high-speed steel, both at 65 Rockwell C, will show some 17 units difference in the Scleroscope test. He gives a conversion table for

Rockwell A, B, C, and D; Brinell; Vickers or Firth and Scleroscope, applicable to tool steels.

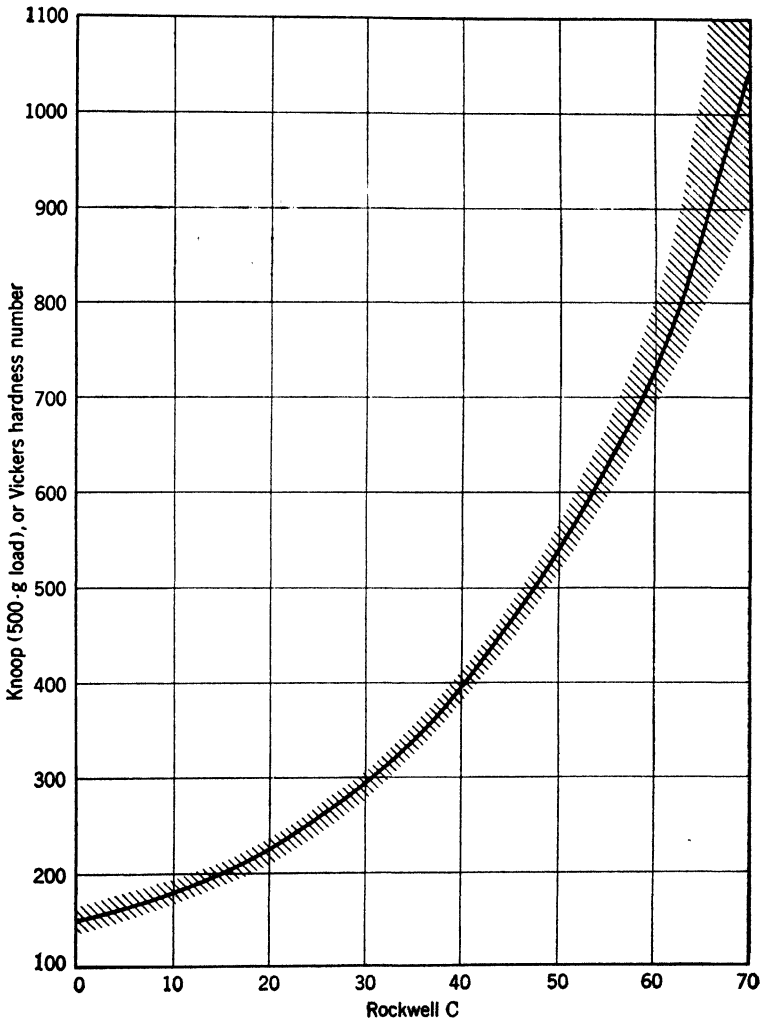


FIG. 87. Conversion chart. (Rockwell C and Vickers or Knoop)

**Hardness versus Tensile Strength.** The indentation hardness values bear a rough relation to the ultimate tensile strength and may hence be used as a rough-and-ready approximation to the tensile values, but not as a sure substitute for a tensile test. Figure 90 shows that reasonably good conversion, within about 10,000 psi can ordi-

narily be had, between 10 and 50 Rockwell C. Up to around 600 Brinell, where the Brinell values become gravely in doubt, the Brinell multiplied by 500 gives a fair approximation of the tensile strength,

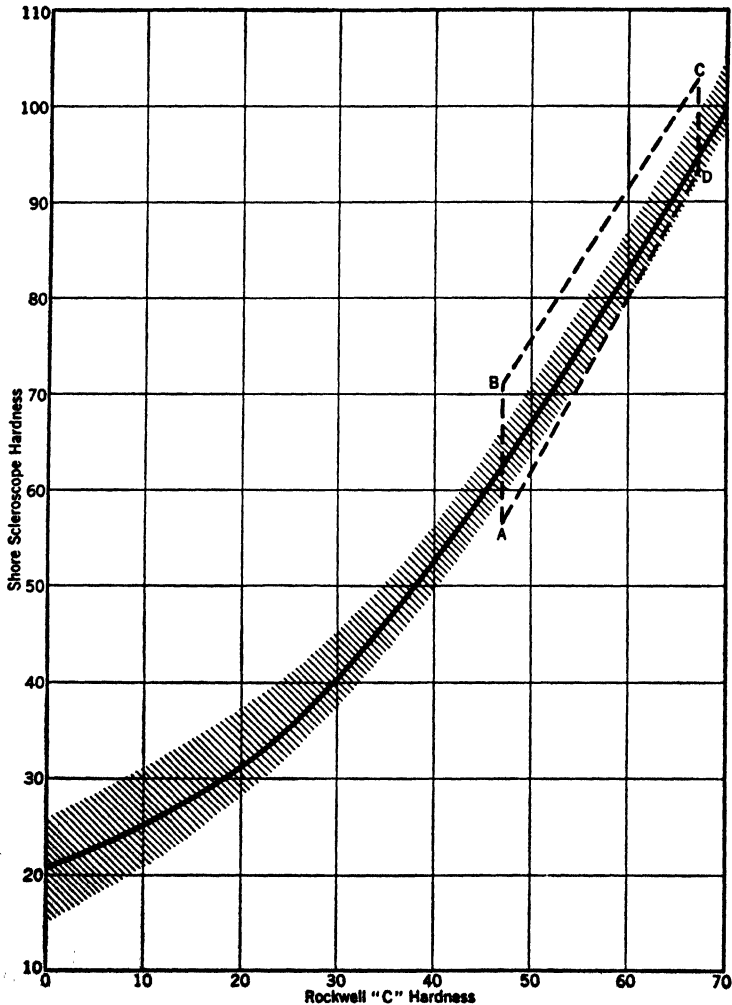


FIG. 88. Conversion chart. (Shore and Rockwell C)

though this conversion is a trifle low below about 200 Brinell. Figure 91 shows the average of a number of conversion tables. The approximation— $500 \times \text{Brinell} = \text{tensile}$ —is ordinarily within the limits of accuracy of the tensile and hardness tests. Using 485 instead of 500 may be more accurate<sup>95</sup> but not so easy to remember.

Conversion curves to tensile from Brinell and Rockwell A, B, and C are shown in Figs. 92 and 93.

Mottley<sup>37</sup> gives a tensile-Brinell curve for the mild-alloy high-tensile steels included in his investigation. The slope of his curve differs slightly from that of the solid line in Fig. 91, emphasizing the

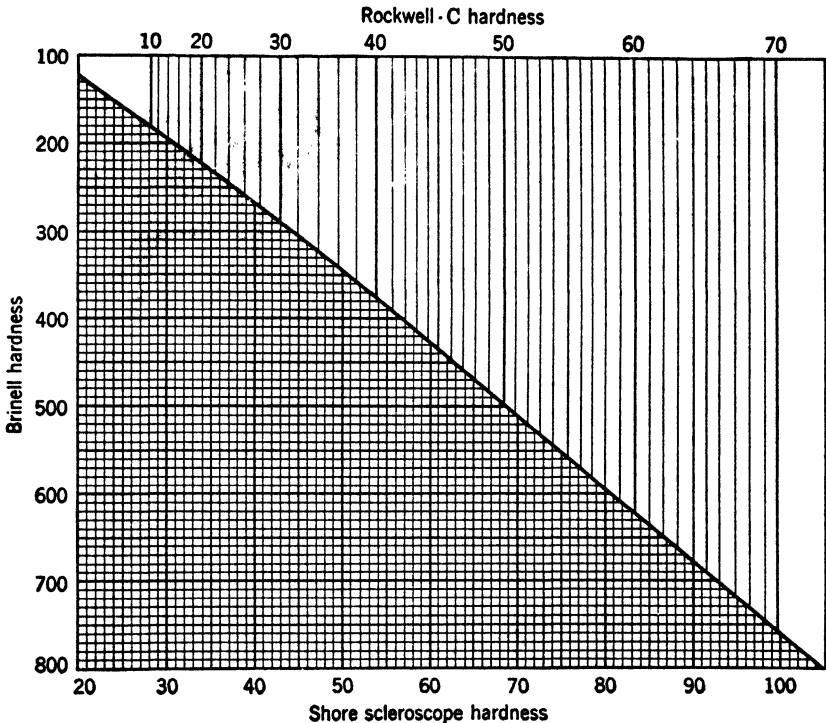


Fig. 89. Conversion chart. (Brinell, Rockwell C, Shore, Kommel<sup>28</sup>)

fact that different "families" of steels tend to be consistent among themselves, but a bit different from another family. Since the yield ratio of the mild-alloy steels is fairly constant, Mottley includes yield strength as well as tensile strength in his plot. Crook<sup>35</sup> gives yield-strength curves or tables for ordinary, low-alloy, and high-alloy steels differently heat-treated. The yield ratio is so dependent on structure that no single formula holds for all. For steel quenched to all-martensite and then tempered, Crook cites Herres for data leading to the formulae: Tensile = 542 (Brinell) - 20,000; yield = 500 (Brinell) - 20,000. Slack-quenched and normalized steels have lower yield ratios.



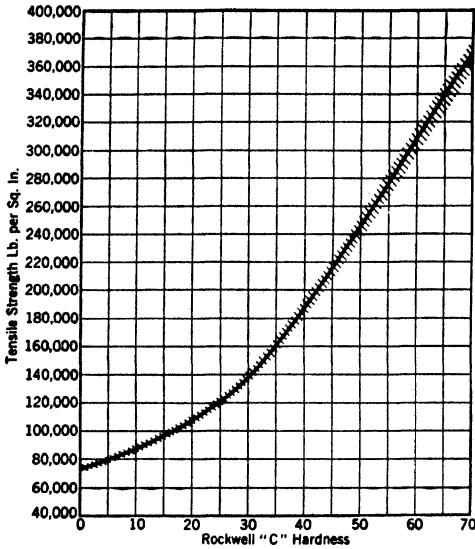


FIG. 90.

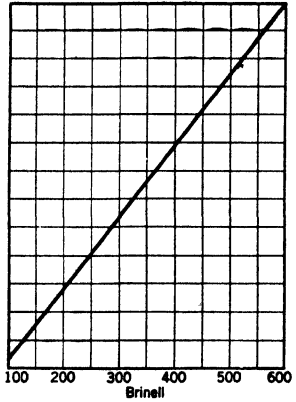


FIG. 91.

Conversion charts. Tensile strength. (Rockwell C and Brinell)

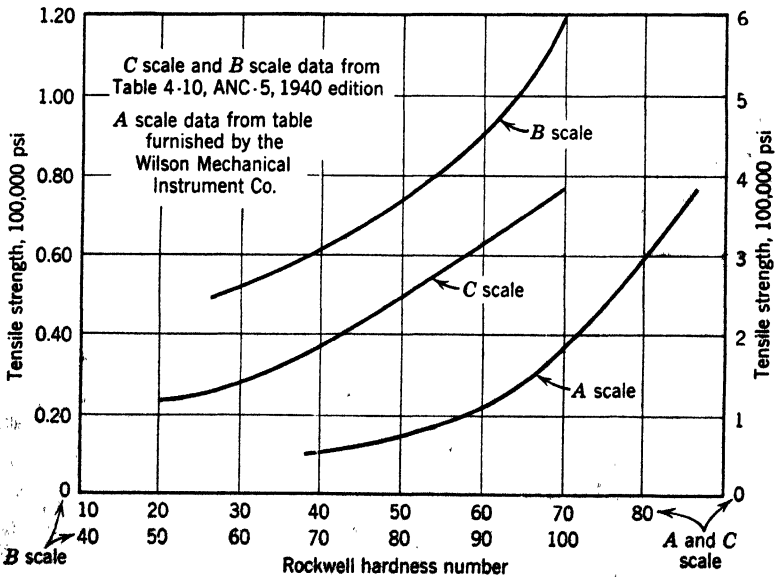


FIG. 92. Curves for the conversion of Rockwell hardness readings to indicated tensile strength. (Rieben<sup>86</sup>)

**Magnetic Hardness.** There is close relationship in steel of the same chemical composition, between certain magnetic properties and the quenched, or quenched and tempered, structure which is reflected in hardness. Thus, an equivalent of hardness testing can be per-

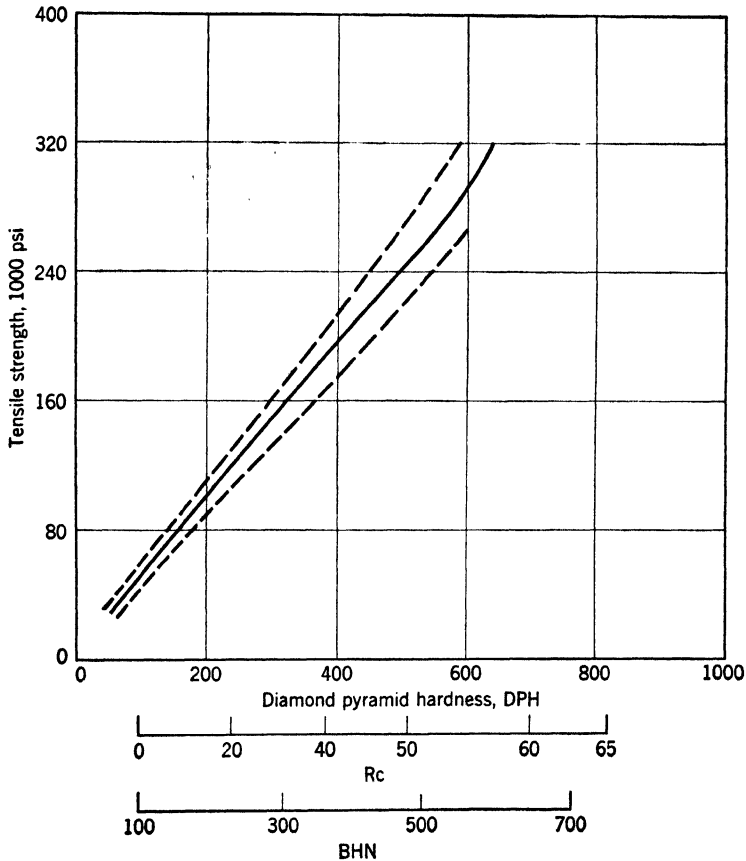


Fig. 93. Approximate relationship between hardness and tensile strength of steels. (Scott and Gray<sup>85</sup>)

formed through magnetic testing. Magnetic testing is nondestructive, without even the slight mar resulting from a hardness indentation.

This offers a speedy method of routine inspection for slack quenching and for differences in tempering. It is usually carried out by reference to a standard of the desired composition, size and shape, structure, internal stress, and hardness. If these attributes are the same in the piece being compared as they are in the standard, the

magnetic behavior of each will be closely alike, so that the method is really an identity test.

When identity in composition, internal stress conditions, or other factors affecting magnetic behavior are not important, the test may be too rigorous, but may even then be used to sort out pieces for further testing by other methods. See discussion under "Magnetic Testing," *ASM Handbook*, and also Cyclograph, p. 209.

Post and Fenstermacher<sup>80</sup> utilize magnetic measurements to determine the proportion of martensite in the cone specimen used for hardenability determinations on shallow hardening steels, or in the Jominy end-quench hardenability specimen. The C content must be known in order for the readings to be interpreted.

## FATIGUE

Since engineering design has been based primarily on yield strength or tensile strength, with some factor of safety, few machines fail from static overloading unless they are involved in an accident which throws upon them a stress far exceeding the design value. However, many failures do occur in parts subject to repeated stress well below the yield stress that can be borne in a single static loading. Indeed, it is often stated that 90% of actual fractures occur owing to repeated stress; and drivers of early automobiles will remember cases of failures of springs and axles, when the part broke off with no sign of deformation, with a fine-grained "detail" fracture over most of the break, and with signs of the ends of the break having rubbed together for a long time before final failure. Perhaps the final fracture was coarse grained in comparison to the detail fracture, from which arose the garage mechanic's erroneous phraseology that the part had "crystallized in service." Such fractures start from a nick or a notch of some sort (external or internal, perhaps a nonmetallic inclusion in the latter case), or a poor fillet that brings about a concentration of stress at some local point. Under long-continued repetition of stress, a crack finally starts at this locally overstressed point and then progresses to final failure.

Much attention has been paid to the problem of fatigue failure, and its nature is now quite well understood. At stresses higher than a certain stress, termed the endurance limit, there is a tendency for the steel to be locally overstressed (but without material plastic flow, so that the over-all dimensions are not affected). Though it is difficult to prove, it is not unreasonable to assume that this spot, because of

some local defect of the notch type, or because of inability of the material to distribute the stress, is stressed above its yield strength and might be considered to become finally so hardened by cold work that it becomes brittle and starts to crack. The damaged spot may be so submicroscopically small that no evidence of its existence can be obtained till the crack actually starts.

In contrast to this damaging effect of overstressing,<sup>86</sup> there may be a helpful effect of understressing. Stresses below the endurance limit tend to produce something analogous to cold working, again without perceptible changes in over-all dimensions, but with a change that is reflected in a perceptible though small increase in hardness and tensile strength.

These two effects, damage by overstress and improvement by understress, come to a balance at an intermediate stress, the endurance limit. This is the maximum stress which the specimen can endure for any number of repetitions, however large.

The endurance limit is of very great importance in design of repeatedly stressed parts. Only in very soft steels is the endurance limit higher than the yield strength. Some such soft steels can be pulled out and pushed back by rapidly applied tensile and compressive stresses that are perfectly balanced, as in a rotating shaft, without showing either permanent deformation or failure from fatigue, though, if the same range of stress were applied with the stresses unbalanced, deformation would ensue.

In any but the softest steels the endurance limit is below the yield stress and bears no direct relation to it. Often a steel will fail under repeated application of loads that are entirely safe from the point of view of static loading. This accounts for the service failures from fatigue when design pays no attention to the endurance limit.

The demands of design made it necessary to evolve methods of fatigue testing.

**Fatigue Testing.** Repeated direct-tension or direct-tension-compression loading has been much used on Al- and Mg-alloy sheet, riveted joints and welds, but relatively few data have been so obtained on steel. Most of the available data have been obtained by the convenient rotary-bending method. Such data can be fairly well translated into terms of direct loading. In rotary bending a shaft is loaded and rotated so that each element of the surface is alternately loaded equally in tension and in compression. Axial loading is also used, when the tension and compression loads may not be equal. However, the average of these unequal loads has about the

same effect as that of fully balanced loads equal to this average, as long as the stress range is not high enough at either end to exceed the yield strength in tension or compression. The main difference in the two types of loading is that axial loading stresses the whole cross section uniformly, whereas rotary bending has the maximum stress at the surface. Hence case-hardened or nitrided steels or those that have been work-hardened on the surface, that is, those with fatigue-resistant surfaces, show higher endurance limits in rotary bending than in axial loading, and the results of rotary bending are not directly applicable to the performance of nonhomogenous steels unless they are thus stressed in service.

Ordinarily, however, the rotary test is adequate to evaluate the endurance properties of a steel. The test is carried out on several types of specimens and machines, machines of the R. R. Moore type being widely used.<sup>39</sup>

Most rotary-beam tests have been run on small machines, using specimens only about  $\frac{1}{4}$  in. in diameter at the breaking section. Mechanical work, introduced in polishing the specimen<sup>40</sup> strengthens the surface, and on a small specimen in which the stress gradient is very steep the strengthening affects the results. Specimens 1 in. in diameter or larger show less of this effect. The data from very small rotating-beam specimens thus tend to be slightly high. Those from 1 in. or larger generally agree well with data from direct-tension-compression tests. Morokovin and Moore<sup>41</sup> discuss possible reasons for the discrepancy in small sections.

In making the test, a series, say not less than four, of like specimens, carefully polished and free from transverse scratches, with the breaking section having a radius of 1 in. or more, is prepared, and the first one is run at such a high load that it fails in a relatively small number of cycles, say, a few hundred thousand. New specimens are run, each at a slightly lower load, until a load is found at which the specimen remains unbroken for at least 5 and preferably 10 million cycles. Experience has shown that in ferrous materials a specimen that will run 10 million cycles unbroken will, with only rare exceptions, continue to run for an indefinitely large number of cycles. This is not necessarily and not usually true for nonferrous materials which should be run to 50 or 100 million cycles and the results reported as "endurance at so many cycles."

Failure may occur after running more than 10 million cycles in very hard steels,<sup>42</sup> and the endurance limit for springs, especially those made from wire, is generally very far below the value one would

expect from its tensile strength.<sup>43-46</sup> The endurance ratio may fall below 0.30. Such behavior is generally ascribed to the presence of internal stress, incipient cracks due to overdrawing, decarburized surfaces, and the greater effect of surface scratches on very hard steels.

For steels, in the ordinary condition however, the 10-million-cycle figure seems justifiably reported as a true endurance limit. Even so, the test is time-consuming and not well adapted for an acceptance test. There is no known satisfactory means of accelerated fatigue testing, though it has long been sought.

The best way to accelerate the test is to build the testing machine in perfect alignment so that it will run smoothly at high speed, and use very high rpm, for example, 10,000, which is now quite common. Another short cut is to run, say, to 5 million cycles only, and then raise the stress, for example, to that corresponding to  $\frac{1}{4}$  million cycles on a virgin bar. A damaged specimen will have a somewhat shorter life. If the specimen was strengthened and not damaged in the original test, it will run *more* than  $\frac{1}{4}$  million, perhaps 1 or 2 million, and this justifies the assumption that it would have run indefinitely at the original stress, the evidence being at least as good as if it had originally been run to 10 million. (Compare the dotted lines in Fig. 97.) The total testing time is shortened by 3 or 4 million cycles.

**S-N Curves.** The number of cycles to fracture at each stress is plotted against the stress, usually on a semilogarithmic scale, as shown in Fig. 94. Thus, stress-versus-number-of-cycles curves are generally abbreviated to "S-N curves." The stress at the value where the curve becomes horizontal is taken as the endurance limit.

If a steel is truly uniform, as the steel in Fig. 94, duplicate runs will give an S-N curve with little scatter, whereas if there is a lack of uniformity the curve will be a wide band. S-N curves would be a very useful evaluation of uniformity were it not so tedious to obtain them.

For no well-understood theoretical reason, it turns out that the endurance limit for steels, and ferrous materials in general, is ordinarily not very far from half the tensile strength. The endurance ratio approximates 0.50. Since the Brinell hardness also happens to have a nearly linear relation to tensile strength, it follows that Brinell hardness is more or less linearly related to endurance limit; if the Brinell value is multiplied by 250, the result approximates the endurance limit.

However, this empirical rule is only approximate, and the upper band in Fig. 95 shows how far from an accurate rule it is. Not only

does the normal value drop below the ratio of 50% of the tensile strength above about 175,000 tensile, but, even below that strength, steels of the same tensile strength may vary 80% in endurance limit, even when the specimens are highly polished and free from surface scratches and notches. This deviation can seldom be traced directly to variations in the usual heat treatment of the steel, though, of

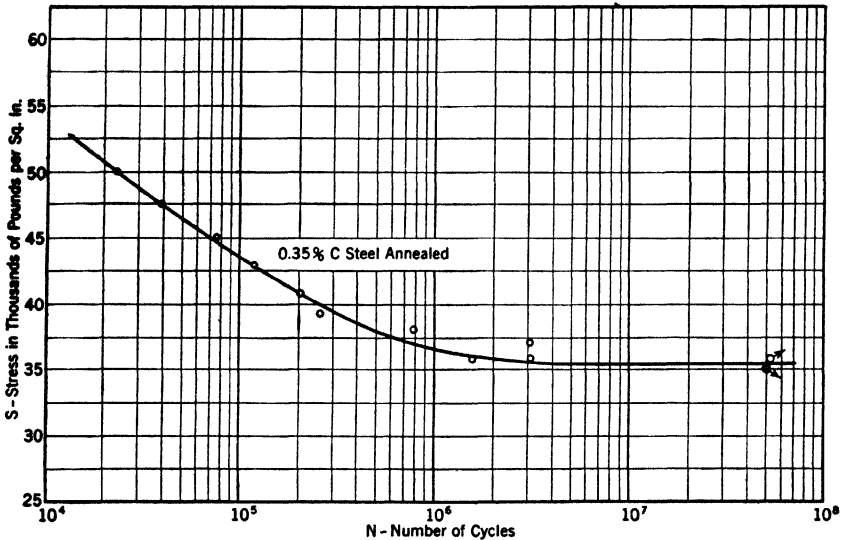


FIG. 94. *S-N* curve for determination of endurance limit. Note the logarithmic plotting for cycles. This is a semilog plot. Log-log plots, with stress as well as cycles plotted on logarithmic scales, are also used. The arrows indicate specimens unbroken after 50 million cycles.

course, the improvement in tensile strength produced by heat treatment is ordinarily directly reflected in an analogous increase in endurance.

However, isolated cases are on record of endurance limits for strong steels given heat treatments aimed to free the steel from internal stress or to prevent internal cracks, which indicate that it should be possible to obtain values that hug the upper boundary of the band for polished specimens in Fig. 95, or even exceed it. Points marked 1 and 2 were obtained by long-continued tempering of oil-quenched Ni-Cr-Mo steels,<sup>47</sup> point marked 3 is for an "austempered" C steel<sup>48</sup> (though the test could not be duplicated), whereas points 5 and 9 represent high-yield-strength low-alloy non-heat-treated steels<sup>49, 50</sup> of high P content of the type to be described in Vol. III.

A case has even been met where one specimen of a certain heat of quenched and tempered steel of 275,000 tensile withstood 210,000 psi in repeated rotary bending for 100 million cycles without failure,

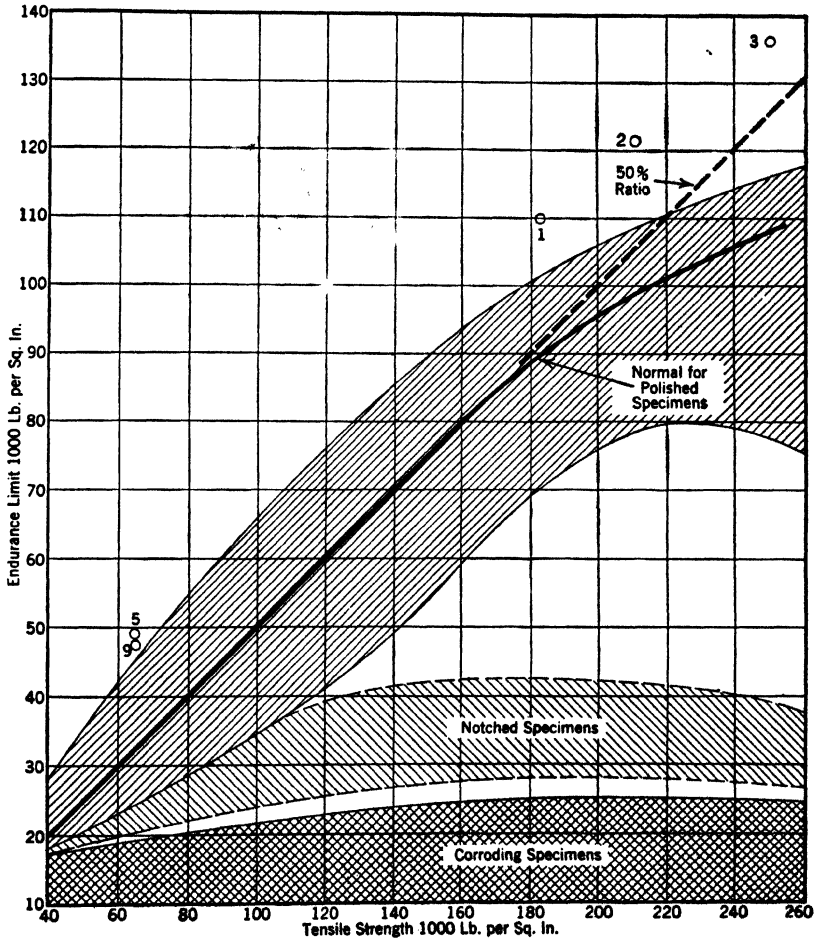


Fig. 95. Relation between tensile strength and endurance limit.

some duplicate specimens giving 20 million cycles unbroken at 150,000 psi; yet another duplicate specimen broke in  $\frac{1}{2}$  million cycles at 105,000 psi, illustrating the wide scatter met in very hard steels. When notched, the endurance limit was 50,000 psi, a high figure.

**Stress Raisers.** Not only do different steels and different lots of similar steels act differently in the standard endurance test with



smoothly polished specimens, but they vary also in behavior when notches are intentionally or unintentionally produced so as to cause local increase in stress.

Even the most notch-resistant steel is severely and very adversely affected by notches. Most fatigue failures in actual parts are directly traceable to some local irregularity such as a toolmark, a number or letter stamped into the steel, a keyway, or a fillet of too small radius. Figures 96 and 97 show how much the endurance is decreased by the

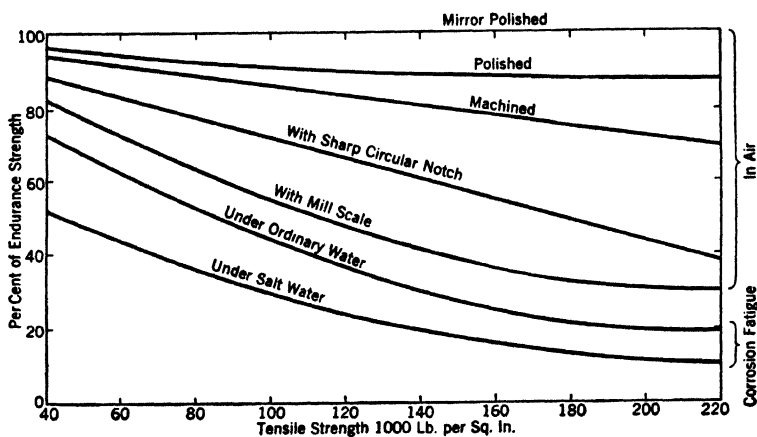


FIG. 96. Reduction of endurance of steel depending on surface conditions. (After Karpov<sup>52</sup>)

presence of a sharp notch. Moore<sup>89</sup> has sketched a number of the common types of stress raisers and indicated the stress increase due to the geometric shape of the piece (Fig. 98). Disregard of stress intensification is a sure way to invite fatigue failures.

**Notches.** The experienced steel treater has a good picture of what features of design are stress raisers, since objects designed with them, when they are quenched, localize the stresses set up in quenching, and the work often cracks. This is still more troublesome because the steel is hard, and, as Fig. 96 after Karpov<sup>52</sup> shows, damage by notches is greatest in hard steels. This figure is drawn on the basis of percentage loss from the endurance shown by polished unnotched bars in air.

Specimens with square notches or V notches, or having changes in section without fillets, give endurance limits ordinarily lying within the band shown in Fig. 95 for notched specimens. Unless notches are avoided, little profit is gained by using very strong steels. Some

austenitic steels which work-harden rapidly, however, are able to withstand the effect of a notch. Russell and Welcker<sup>53</sup> report such behavior on 18 Cr, 8% Ni stainless steel. Cast iron, also, is less sensi

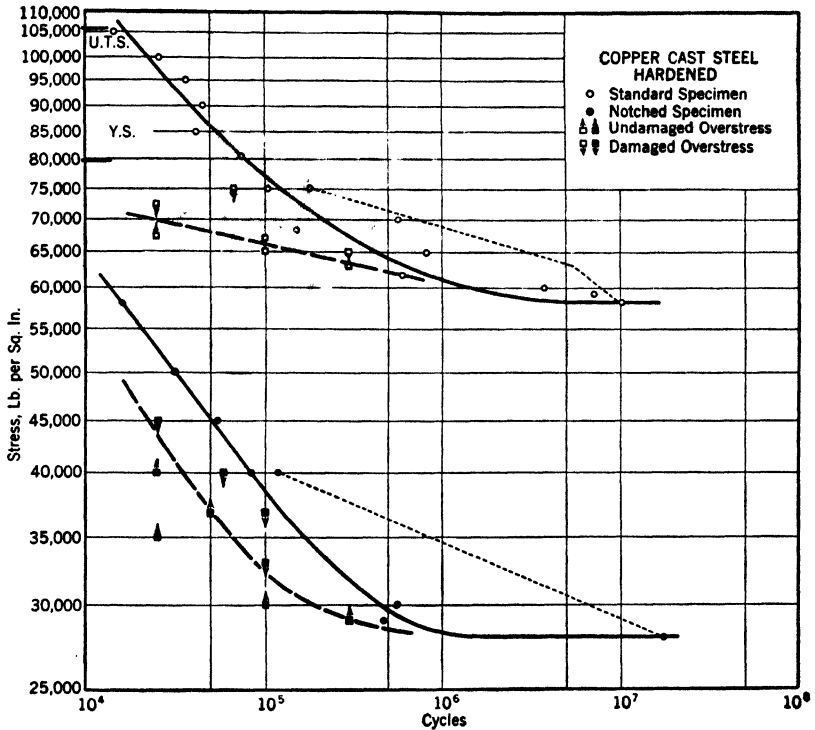
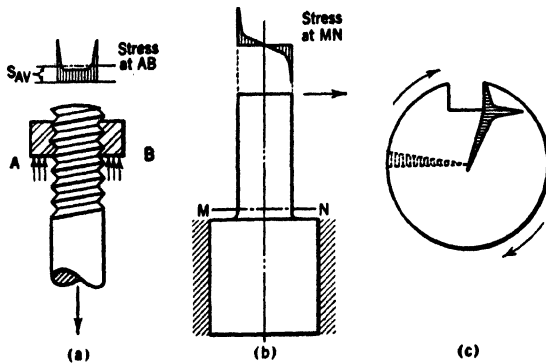


FIG. 97. Log-log  $S-N$  curve for cast 0.23 C, 0.38 Si, 0.61 Mn, 1.20% Cu steel 1650°, 1 hr, air; 930°, 3 hr. Tensile, 106,000; yield, 79,500; 24% elong., 48.5% R.A. Upper curve, polished specimens. Lower curve, 90° square (Kommers<sup>86</sup>) notch, stress calculated on diameter at base of notch. Dashed lines divide the fields for no damage (upon retesting at the endurance limit of a virgin bar) by overstress, of intensity and number of cycles plotted, and for damage by such overstress. The dashed lines are "damage lines." The dotted lines connect the test of a virgin bar that was strengthened by understressing with its retest at a higher stress. (Russell and Welcker<sup>53</sup>)

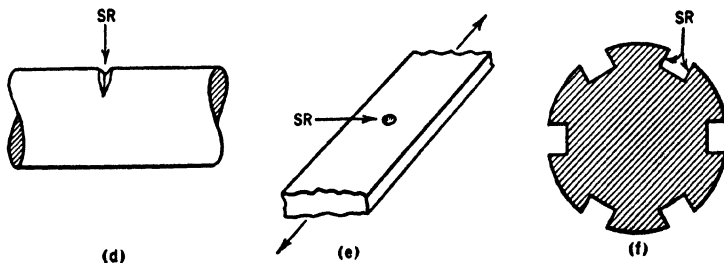
tive to notches than are the ordinary steels. Figure 95 is drawn to represent the ordinary types of steels.

**Factors in Fatigue.** The following points should be kept in mind:

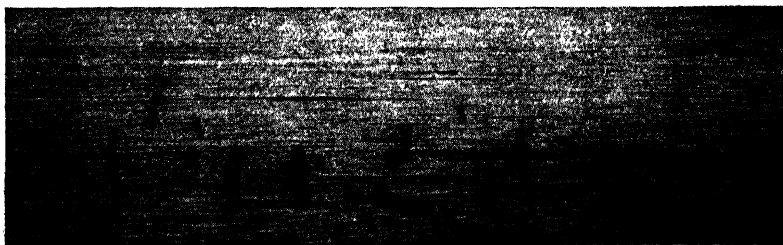
1. Screw threads, grooves, sharp shoulders, nicks, and even tool marks cause high localized stresses and markedly lower the fatigue strength of any steel machine part.



Typical stress raisers in machine and structural parts. (Moore <sup>51</sup>)



Imposed stress raisers. (Moore <sup>51</sup>)



(g) Inherent stress raisers. Shatter cracks in a new rail. (Quick, Bureau of Standards)

FIG. 98. Stress raisers that lead to fatigue failures under overstress, such as screw threads, sharp fillets, key-ways, nicks, holes, sharp corners, cracks. Surface seams and large sharp-cornered nonmetallic inclusions fall into the same category.

2. In figuring allowable stresses under repeated stress it must be remembered that stresses above the yield strength of the metal will cause *elastic* failure. The machine designer and the user of steel parts must be on guard against both elastic failure and fatigue failure.

3. Fatigue strength obtained in tests of small specimens, especially of heat-treated steel, cannot often be obtained in large pieces of steel. It is as necessary to use a "factor of safety" in considering fatigue strength as it is in considering static strength.

But sometimes a part cannot be designed without severe stress intensification; so the ability of different steels or of the same steel differently heat-treated to withstand the effects of stress raisers enters into the choice of the steels or treatments.

**The Endurance Limit and Design.** When and only when the true maximum local repeated stress at the point of greatest stress concentration is known, is the endurance limit usable as a basis for design for freedom from fatigue failure. The difficulty lies in the measurement of that stress.

**Determination of Stress Distribution.** If a scale model of a steel part is made from a transparent material like glass or celluloid, loaded, and examined by polarized light, no color bands appear if the stress is distributed uniformly. Localization of stress produces colored bands whose number is a function of the stress and whose location shows the position of the stress intensification. By such means the evil effect of notches, poor fillets, etc., may be visually demonstrated. This method of study is useful in the design of objects in which the stress concentration cannot be readily or accurately calculated.

Examples of the use of this method are found in the articles by Frocht<sup>54</sup> whereas Peterson and Wahl<sup>55</sup> link together stress concentration and behavior in fatigue, as do Buckwalter and Horger.<sup>56</sup>

**Brittle Coatings.** Putting a thin coating of a brittle lacquer, such as the commercial "stress coat"<sup>57, 84, 85</sup> on the part and subjecting the part to service stresses, under suitable conditions, develop cracks in the coating at the locations of stress concentration. By comparison with the cracking of the coating upon a bar subjected to known stress, a rough evaluation can be had of the maximum load produced by stress concentration in the part.

**Strain-Gage Measurement.** Sensitive electrical strain gages, which quantitatively measure strain over extremely short gage lengths, became highly developed for study of stresses of aircraft during the

war. These, when applicable, afford a more accurate measurement of the maximum stress at a location of stress concentration. A large literature is growing up in this field, chiefly recorded in the Proceedings of the Society for Experimental Stress Analysis.\*

**Use of Local Stress Measurement.** Gadd, Zmuda, and Ochiltree<sup>58</sup> have reported on simulated service endurance tests of actual crankshafts, connecting rods, etc., made from 4340, quenched and tempered to 300–340 Brinell (say 150,000 tensile) and well polished. The maximum local stress on the actual parts was estimated by use of brittle lacquer and by strain gage measurements on  $\frac{1}{8}$ -in. or  $\frac{1}{16}$ -in. spans.

The rotating-beam endurance limit on 1-in.-diameter polished bars was 72,000 psi, in agreement with the average line of Fig. 95. When the actual parts were subjected to endurance tests, there were no fatigue failures when the maximum estimated stress was 61,500 to 70,000 psi, depending on the part used. Considering the approximate nature of even such stress measurements and the usual scatter of fatigue tests, the agreement is not bad. The authors consider this steel at this hardness level representative of a rather high degree of notch sensitivity. Peterson<sup>59</sup> and Boyvey<sup>60</sup> bring out an analogous point of view. Almen and Boegehold,<sup>61</sup> however, concluded that, in the case of rear axle gears, factors of rigidity, scuffing, etc., entered in, as well as fatigue behavior, making difficult the setting up of a simulated service test, and so testing of complete assemblies, especially to bring in the rigidity factor, had to be resorted to.

**Design for Limited Life.** Steel has a definite endurance limit. Many other alloys, notably those of Al, do not; the *S-N* curve never becomes entirely horizontal. Moreover, it is often necessary, especially in aircraft design, to save weight and accept a definitely limited life rather than an indefinite or infinite life. That is, such design deals with the sloping part of the *S-N* curve, at stresses above the endurance limit. Several difficulties arise in such a case, notably that the curve is not a line, but a scatter band, and many duplicate tests on many heats of steel would have to be run to establish the width of the band. The type of specimen, the testing machine, and the testing technique used affect the results. All the specimens, machines, and techniques may agree beautifully as to the endurance limit, but give *S-N* curves of different slopes. Hence simulated service testing of the actual part becomes essential.

\* Central Square Station, Post Office Box 169, Cambridge 39, Mass.

Almen <sup>62</sup> has plotted, in Fig. 99, data obtained by different investigators on the same lot of heat-treated 4340. The plot is logarithmically scaled in per cent of the ultimate tensile. The endurance limits found by various techniques, for polished bars, were all  $80,000 \pm 5000$  psi, but the lives at stresses above the endurance limit were much more widely variant.

The difficulty that the designer finds in applying conventional tests to some of the practical problems of this type is illustrated by failures

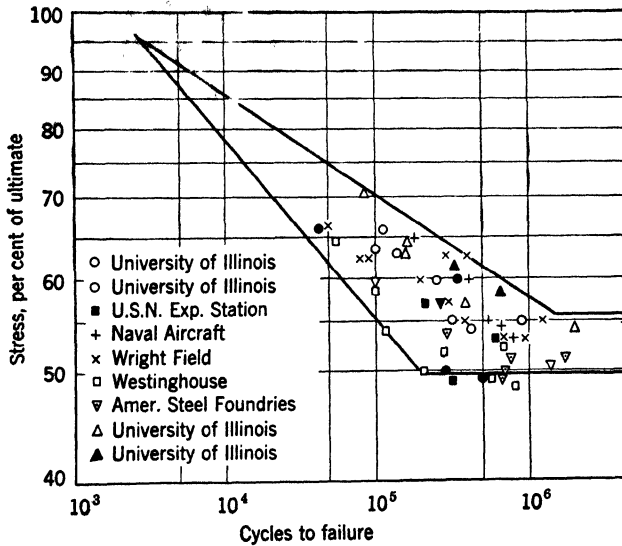


FIG. 99. Scatter of results at loads above endurance limits.

in aircraft landing gears. McLean <sup>63</sup> has discussed these in detail. Failures of a torque fitting were occurring in 120 to 1200 landings, which were obviously fatigue failures, whereas some 6000 landings would be made during the life of the rest of the plane. On account of weight, the design could not be such as to bring the stress below the endurance limit; it had to be on the basis of a 6000 landings minimum life.

Both the actual and the desired life are far up on the high stress end of the  $S-N$  curve. Even when the service stresses in the actual part, roughly determined by stress coat and strain gages, were known, it was not possible to say what the life would be when the stresses were reduced by modifying the design. The measurements did show clearly where the worst stress concentrations occurred, thus

pointing out the places to be strengthened or to have their loading reduced.

To guide the redesign and to prove that the final redesign would be effective, it was necessary to use the full-sized actual parts and rig up special devices for applying repeated loads.

It happened that the design of the torque fitting in use was such that an additional set of torque scissors could be applied which reduced the stress so that a life of some 28,000 to 59,000 cycles was achieved.

It was found then that another part was failing, sometimes in as low as 600 landings. Addition of a reinforcing bracket and reinforcement of a brace fitting to relieve the stresses at the locations of worst concentration brought the life of that part, in simulated-service tests, to around 9000 cycles of what appeared to be the maximum service stress. With these modifications, service failures of the landing gear ceased to be a problem.

No conventional test gave the designer the information he needed in this instance.

A further design difficulty is lack of knowledge of the magnitude and number of applications of the maximum stress that will actually be met in service, a matter particularly difficult in aircraft design. Vibration may set up resonant stresses which cannot be duplicated in anything but the actual aircraft. A notorious case of this was in the Graf Zeppelin. The engines behaved well on the test stand, but their crankshafts failed in fatigue in flight. The trouble was traced to resonant vibration.

The intensity of gusts of wind and the stresses imposed in landing are not readily predictable; nor is it known in what order such exceptional loads will be interspersed by periods of understressing which might bring about improvement and repair some of the damage done by overstressing.

Such factors make the prediction of the fatigue life of an actual structure in actual service very difficult. It is easy to impose a single static proof load and see whether the yield strength has been exceeded, but far from easy to explore experimentally the behavior between the endurance limit and the yield strength under repeated loadings that duplicates service.

Miner<sup>64</sup> discusses one method of approach to this problem. The principles would be equally applicable to design of steel parts for use at stresses above the endurance limit. He emphasizes the necessity for experimental determination of the  $S-N$  curve on specimens

or, still better, actual parts that have the exact stress raiser that will be contended with in practice.

If there were some easy way of inspecting an actual part that has seen service and determining whether it has been damaged and should be replaced by a new part, or whether it is fit for further service, much of this uncertainty would be removed. Unfortunately, little is known about the fundamentals of damage, and so this approach is not yet feasible. Enough has been done to indicate that some materials, even some with mediocre endurance limits, have exceptional ability to withstand overload without rapid damage, while others are especially vulnerable. It is a tedious matter, especially so because of the scatter among duplicate fatigue tests, to examine the fundamentals. Much effort has been expended in the hope that X-ray methods would detect submicroscopic changes, without avail. A possibility still exists that magnetic "identity" tests, in which an unused part is compared with a used one, may have applicability but even this does not seem very hopeful. Cavanagh<sup>87</sup> is optimistic about such tests as applied to wire and wire rope. The only well-tried inspection method at hand is visual examination or magnifying the part to reveal actual cracks, but the stage of crack development is far too late in the history of damage to be a true solution to the problem.

Although study of the propensity of a material to fatigue damage is far from being a conventional test, the topic needs mention here.

To perform well under stress-raising conditions, it is generally believed that high ability for local deformation, such as that evidenced by high reduction of area in the tensile test, is helpful in allowing the material to yield locally at the base of a notch and thus redistribute the stress and lower its intensity. However, the materials of high reduction of area are not always the best in notched fatigue tests. Not only is there required ability to redistribute stress, but also ability to resist the propagation of a crack once it is formed. The quality needed has been termed by H. F. Moore "crackless plasticity,"<sup>88</sup> but how it is to be measured and evaluated is not yet certain.

In parts designed with reasonable attention to freedom from stress raisers and to the endurance limit as determined on polished specimens, fatigue failures in service seem to come mainly from the imposition of a few accidental overloads; so the ability of a steel to withstand overload without damage counts perhaps as much as its



endurance limit on a polished specimen. We need a test to evaluate that ability.

**Damage-Line Testing.** The test consists in subjecting bars to a known number of cycles of known overstress and then retesting at the endurance limit for virgin nonoverstressed bars, to see whether they have been damaged or whether they will run the usual 10 million cycles without evidence of damage in the prior overstress run. From such tests a "damage" line can be drawn, as French,<sup>42</sup> and as Russell and Welcker<sup>53</sup> have done (see Fig. 97). Retesting could also be done at stresses above the endurance limit. Such work is slowly throwing light on the properties needed to combat fatigue failures in service, though it is not yet possible to correlate them closely with properties determinable in static tests or in ordinary endurance tests.

**"Save the Surface."** In most types of service under repeated stress, the maximum stress applied is at the surface. If the surface skin is weak, fatigue failure is invited. The evil effects of a decarburized skin on a spring are common knowledge and have been shown in detail by Hankins and Becker<sup>66</sup> and by Austin.<sup>67</sup> Methods for hardening without decarburization are demanded, and the susceptibility to, or freedom from, decarburization is an important factor in the choice of an alloy steel.

Conversely, if the surface is made stronger, as by carburizing, nitriding, or shot peening the endurance will be raised in a rotary-bending test or in service where the skin takes the maximum stress. Very great increases in endurance limit, for example, are produced by nitriding, and considerable increases by cold-working the surface. However, if the stress is applied axially (as in a connecting rod, for example), this improvement, as shown by rotating-beam tests, may be illusory since failure may start beneath the surface. A hard but brittle surface layer in which a crack can readily be formed may be dangerous because the crack may readily propagate through the softer core. A case in point is the galvanized heat-treated bridge wire discussed by Swanger and Wohlgemuth;<sup>68</sup> other cases have been described.

**Shot Peening.** Cold-working the surface, as by rolling with suitable tools, more commonly by shot peening, raises its endurance limit and induces compressive stress at the surface. Even the mild cold working involved in the usual polishing of a fatigue specimen has a materially beneficial effect. Great improvement has been brought about in the service life of springs by shot peening. Almen has developed the technique of control of shot peening so as to avoid over-

cold-working and the production of tiny surface cracks, and has published widely on the topic.<sup>82</sup> In terms of life at a stress chosen around the knee of the  $S-N$  curve, the improvement may be most spectacular, whereas, if the endurance limits of peened and unpeened materials are compared, the improvement, though real, is far less imposing. Moreover, in some cases (a small minority) where marked improvement would ordinarily be expected, little or no improvement is met. The method is not a blanket cure-all for every case of fatigue failure, but it is so often and so highly effective that it deserves major consideration. Jensen and Moore<sup>88</sup> cite a case where the stress introduced by shot peening would hardly be expected to be helpful, but actually proved to be so.

**Local Cold Work.** It is difficult to visualize the localized cold work that is postulated to produce damage and ultimate cracking. Our ordinary concept of ductility leads us to expect stretching under load over a considerable "gage length," as a precursor to necking and final fracture, even though, in the plastic action which redistributes static stress in, for example, a riveted structure, it is recognized to be quite local. For such stretching and redistribution of load, the stress has to rise above the static yield strength. Quite a different condition exists when the stress is well below the static yield strength but above the endurance limit.

A case in point is that of spot welds in high-strength aluminum alloys.<sup>96</sup> There is a tiny zone of softened material about the weld, produced by the heat of welding, which does not very seriously affect the strength in static tests. Tests on pieces joined by a single spot weld likewise show fatigue strength somewhat but not too far below expectations based on the static tests. But, when pieces are tested in fatigue that are joined by multiple spots, shockingly low results are obtained, and these low results are corroborated by service experience. Some one of the multiple spots carries an excessive load and is vulnerable to fatigue damage. A crack starts at that spot. Probably a prestressing by a single high overload which would deform enough material to redistribute the stress and make all the spots share it equally would be helpful. But, when the vulnerable spot is forced to take the unshared overstress repeatedly and the area undergoing cold work is very limited, over cold working and ultimate cracking can ensue, just as in shot peening.

Materials of vast ductility, and of high endurance limit when conventionally tested, are not immune from such damage. Austenitic manganese steel is such a material. Caille<sup>97</sup> has reported that pins

of Hadfield steel in chain links develop a locally work-hardened surface, which cracks, and the cracks propagate into fatigue failures. Yet conventional notched fatigue tests of austenitic 18:8 would lead one to expect immunity in so ductile and work-hardenable a material as an austenitic steel, whereas the indications are that a multiple-spot-welded joint in an austenitic steel may have a very low order of fatigue resistance. Nothing seems to have been published on this, but all the indications point to the wisdom of applying the maxim,<sup>98</sup> "Endurance tests on finished parts are the only sound basis for design."

**Corrosion Fatigue.** Finally, there should be mentioned the further fact that, since steels are sensitive to the effect of notches, the notches caused by corrosion are highly detrimental, and that, moreover, *simultaneous* repeated stress and corrosion are even worse than corrosion followed by repeated stress. The "corrosion-fatigue" phenomena, originally brought out by Haigh<sup>69, 70</sup> and emphasized by McAdam,<sup>71, 72</sup> have to be considered in any service involving the two factors.<sup>73</sup> Sucker rods in oil-well service are an outstanding example.<sup>74, 75</sup>

Under combined corrosion and repeated stress there is no true endurance limit, and the endurance of a very strong but corrodible steel is very little if any greater than that of a much weaker equally corrodible steel. This is indicated in the lower part of Fig. 95 which shows that under active corrosive conditions the endurance of all corrodible steels falls practically to the same level. Under such conditions strengthening by heat treatment is unavailing.

There is no true corrosion-fatigue limit, the stress-cycle curve tends to be a straight line instead of flattening off, although in Figs. 95 and 96 the corrosion-fatigue results have been plotted as if they were endurance limits. If fatigue tests are carried out under corrosive conditions, a corrodible steel will ultimately fail, no matter how low the load, though failure may be long delayed. In testing, the test cannot be speeded up, because time must be allowed for corrosion to take place. The behavior under combined corrosion and repeated stress indicated in Figs. 95 and 96 is for ordinary corrodible steels. If an alloy steel is completely incorrodible under the corrosive conditions to which it is subjected, it will show the normal *S-N* curve, that is, the same as tests in air. Even the most corrosion-resistant steels are seldom so completely resistant as to show no loss of endurance in corrosion-fatigue tests.

All these effects of surface condition and the other factors affecting life under repeated stress are more fully dealt with elsewhere.<sup>76</sup>

**Corrosion Testing.** Corrosion by itself may not be a major problem in most services for which heat-treated steels are required, for the steels used in moving parts are generally automatically coated with oil, and steels subject only to static stress may usually be painted or otherwise protected. When the necessity does arise for corrosion testing, the important thing is to see that the test conditions duplicate, or are sufficiently analogous to, the conditions of service. No "standard" test for corrosion exists, because corrosion resistance is not a property inherent to the steel, exclusive of its service environment.<sup>77-78</sup> No marked difference exists in the corrosion resistance of C steels—they all rust; hence further attention to corrosion is deferred until highly alloyed steels are discussed in Vol. III.

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## CHAPTER 7

### NOTCHED-BAR IMPACT TESTING

We have seen that among the values determined in the conventional tensile test, only yield strength is directly applicable to design. The value for fatigue strength is applicable only when the designer knows what the actual stresses are at points of stress concentration. Both these values are in pounds per square inch. We now come to another conventional test, one that gives quantitative values, but values that are of no direct application whatever to design, because they cannot be put into pounds per square inch. This is the notched-bar impact test, used as a purely qualitative index of "toughness." Because of its qualitative nature it is not properly classed with the tests mentioned in Chapter 6. The discussion under fatigue has brought out the marked intensification and localization of stress at the base of a notch even when the applied stress is low. When a notched bar is broken by a single application of applied stress, very great differences appear between steels of different structures, even when they show the same tensile and yield strengths and the same ductility in the tensile test. Some are tough and deform freely; others snap off almost like glass.

For instance, Sims<sup>1</sup> tells of two bars of cast steel from the same heat which were heat-treated differently. These were tested by the tensile test and the notched-bar impact test. Though the results in the tensile test were nearly identical, they were materially different in the impact test, as the data in Table 6 show.

TABLE 6

	<i>Heat Treatment</i> A	<i>Heat Treatment</i> B
Tensile	98,250	101,500
Yield	63,000	66,000
Elong. %	27.0	27.5
R.A. %	55.5	57.5
Izod	18.5	48.0

Bar *A* was normalized from a rather high temperature after a long period of heating, resulting in a coarse-grained structure; bar *B* was normalized from just above the critical temperature, resulting in a fine-grained structure. One cannot say that one of the two treatments is right and the other wrong without knowing the type of service to which the steel is to be put. In high-temperature service the coarse-grained material with the lower impact resistance should be the better, whereas it might be the poorer in shock-resistant service at ordinary temperature.

The impact test shows *A* to be exhibiting some approach to relatively brittle behavior, and *B* to ductile behavior. These results do not tell us that condition *A* is always brittle and *B* always ductile. They only mean that these things are true insofar as the particular set of loading conditions imposed by the test is concerned. Some types of loading conditions could probably be found in which a steel with the properties of *B* would act no more ductile than one with the properties of *A*.

In other words, most steels have a certain propensity toward brittle behavior, which varies from steel to steel. This difference has some importance in relation to service involving sudden shocks of high intensity, and impact resistance may govern the choice of material in such service. However, many materials that are almost glass-brittle under impact in the presence of a sharp notch act tough when they are formed with fillets of reasonably large radius, such as good engineering design always calls for. Under the tensile-impact test, which uses an unnotched bar but applies the tensile stress by a sudden pull, they may show quite as good ductility as in a static tensile test; so the suddenness of the blow, up to ordinarily high rates of loading, is not primarily what makes the steel act brittle.

On the other hand, at extremely high rates of loading, as during the firing of a big gun or when a projectile is fired against armor, it is thought that steels that act alike at lower velocities may act differently and that brittle behavior may appear that would not be expected on the basis of static tests.

There is a marked size effect. A steel that shows toughness in all small-scale tests may, and usually will, behave in brittle fashion in a very large section.

The appearance of brittleness is largely a function of the loading conditions and not wholly inherent in the steel. Steels do vary greatly in their behavior under the same loading conditions, but it would probably be possible to find conditions of loading under which

almost any ferritic steel could be made to fail in a brittle fashion, though fortunately such conditions are very rare in service.

Most of the phenomena concerned are brought out even on a slow-bend test, like the Humphrey,<sup>2, 32</sup> but for speed and convenience the bars are generally broken by impact.

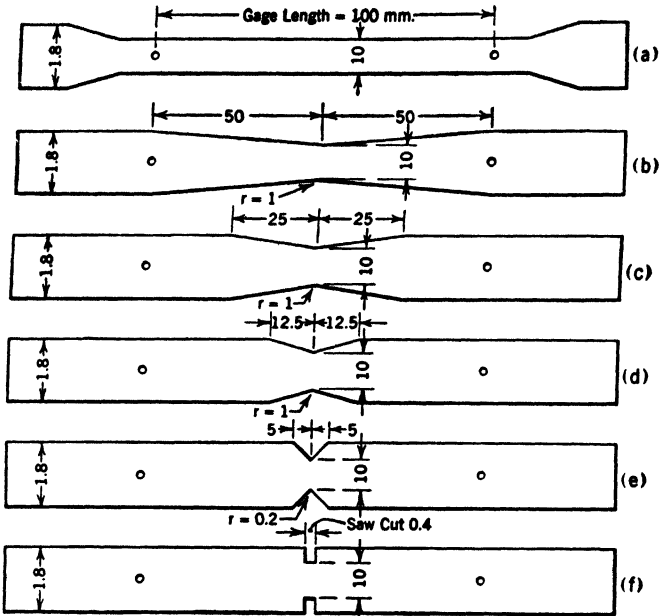


FIG. 100. Notched cylindrical tensile test bars, in section (dimensions are in millimeters).

The notch effect in tensile tests, on bars shown above:

Bar	Tensile	Elong., %	R.A., %	Effective Elong.*	Energy absorbed, (kg-cm)
a	56,000	35.7	69.0	233	10,147
b	62,000	7.5	64.6	182.5	2,281
c	65,500	5.6	58.2	139	1,772
d	73,500	3.6	42.7	74.5	1,232
e	84,000	2.3	32.8	49	905
f	91,500	1.5	22.5	29.1	710

\*  $\frac{R.A.}{1 - R.A.}$ , estimated effective % elongation. (Ludwik and Scheu, quoted by Hoyt<sup>3</sup>)

The differences in propensity toward brittle behavior are sought out and accentuated when the volume that can take part in the deformation is limited, that is, when the energy of fracture must be absorbed in a very small volume. For example, if a tensile test is made on bars of the same steel with and without notches, the cross section at the base of the notch being the same as that of the un-

notched bar, the tensile strength goes up and the elongation goes down in the notched bar. The energy absorbed drops rapidly, as Fig. 100 and the accompanying table show.<sup>8</sup>

Obviously, a steel with high reduction of area in the regular tensile test tends to give a good account of itself under such conditions, but the correlation is not perfect. Increases in tensile strength,

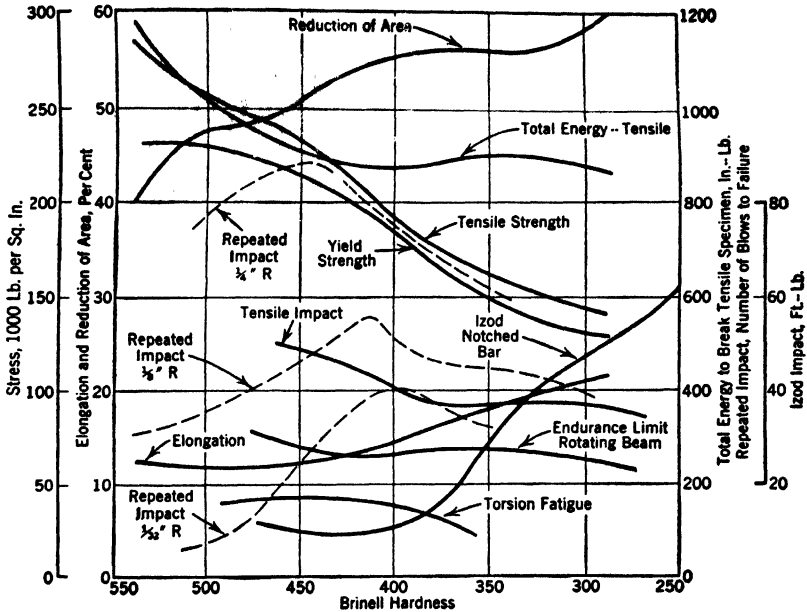


FIG. 101. Mechanical properties of heat-treated 3140 steel, plotted in relation to hardness. (After Knowlton <sup>4</sup>)

Brinell hardness, yield strength, energy absorbed in the tensile test, or endurance limit are not reflected in an improved notch-impact value. On the contrary, Fig. 101 indicates, as is well known by experience, that great strength or hardness can be obtained only at the sacrifice of toughness as measured by the notch-impact test.

**Machines and Specimens for Impact Testing.** In order to bring out the propensity toward brittle behavior, standard tests have been devised in which a notched bar is fractured by bending, under a single impact. The two types of tests most used are the Izod and the Charpy. The energy absorbed in fracture is measured, as by attaching the striking hammer to a pendulum, which is allowed to swing from a known angle, to hit the specimen at the low point of

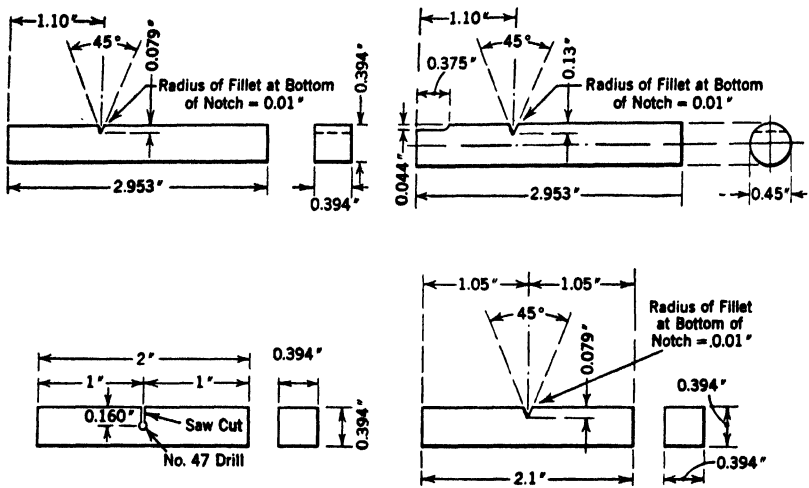


FIG. 102. Some of the forms of notched-bar impact specimens. Keyhole Charpy bar (preferred type), bottom, left. Alternative Charpy bar, bottom, right. Square Izod (preferred type), top left. Alternative round Izod, top right. With these various specimens in vogue, it is evident that a "Charpy" or "Izod" figure reported by different investigators does not necessarily relate to the same bar. Unless it is known that the same bar was used, data from different investigators may not be comparable.

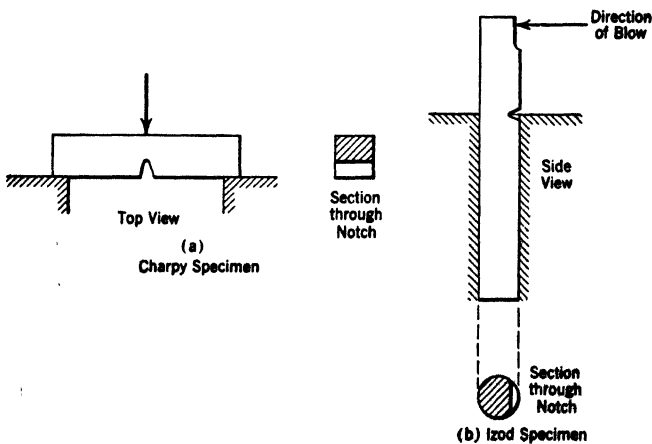


FIG. 103. Method of loading single-blow notched-bar impact specimens.

its swing, and to rise, in the remainder of its swing, to a smaller angle than it would have risen to had it not given up part of its energy in breaking the bar. From the angle to which it actually rises, the energy absorbed by the bar may be computed.

In the Izod test the bar is vertical and rigidly gripped at the lower end. The hammer hits above the notch and breaks the bar by cantilever action. In the Charpy test the bar rests at each end against rigid supports, the notch being in the middle and the hammer striking at the notch. The bar is thus broken in transverse bending (see Fig. 103).

The Charpy bar is square in cross section so that it may rest on a flat side against the supports. The notch for the Charpy is ordinarily made by drilling a small hole tangent to the center line of the bar and cutting down to that hole with a saw cut. Alternatively, the notch may be made as a V.

The Izod bar uses the V notch and is usually square like the Charpy, but may be round, with a flat milled on it to receive the striking hammer. The several bars are shown in Fig. 102.

The  $10 \times 10$ -mm ( $0.394 \times 0.394$ -in.) Charpy bar with a keyhole notch 5 mm deep, leaving  $10 \text{ mm} \times 5 \text{ mm}$  ( $0.394 \times 0.197$  in.) back of the notch, has been adopted in most of the Continental countries.<sup>5</sup>

Results are expressed in meter-kilograms or foot-pounds of energy absorbed for a particular bar, the notch and the dimensions of the bar being stated. In the United States, results of either test are normally expressed in foot-pounds. In foreign literature, impact results are sometimes expressed in meter-kilograms per square centimeter; that is, the energy absorbed is divided by the area back of the notch. Even when expressed in such units, results of tests made with different sizes of bars and different notches are *not* interconvertible.

**The Notches.** The *ASM Handbook* states that the general opinion is that the keyhole notch is superior to the V notch for general testing. On this basis the Charpy test, in which the commonly used bar has the keyhole notch, would be the logical one to select as standard. Unfortunately, in some industries, the available data are chiefly on Izod bars, and specifications call for Izod tests, so that Izod rather than Charpy data continue to be accumulated. This is the case with steel castings. The Izod bar is often made with several notches, on different faces, the bar being regripped after the fracture of each section. Less stock and less machining costs are required for duplicate determinations than when Charpy specimens are used.

There is usually no material advantage in using one of the two types of test as compared with the other, since both usually tell similar stories. However, impact tests at high and low temperatures are becoming increasingly important, and the Charpy bar can be heated

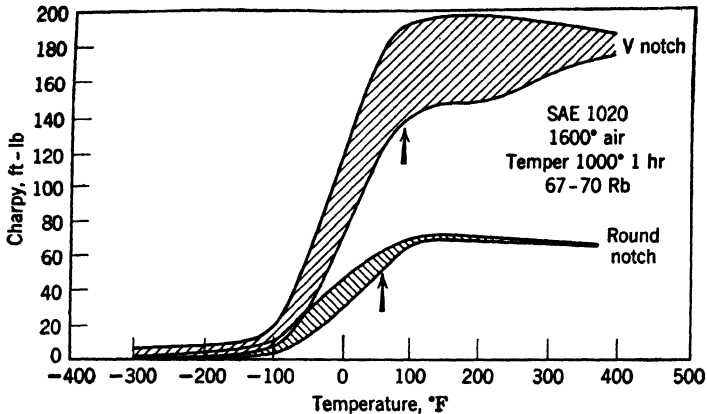


FIG. 104. V notch versus round notch results on a soft normalized steel.<sup>34</sup>

or cooled to the desired temperature, laid on the supports, and broken without delay; whereas the Izod has to be clamped into the base, which takes time, and the base also has to be heated or cooled to the test temperature. Even when one is not primarily aiming at

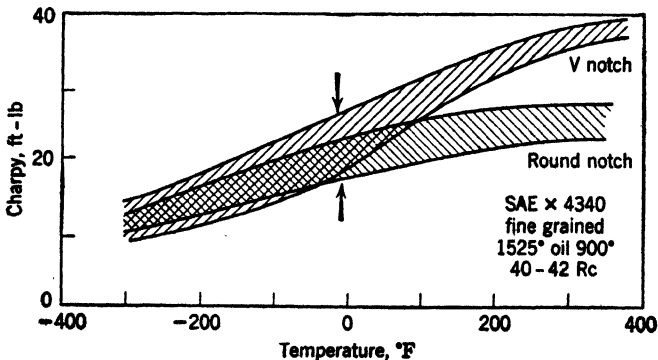


FIG. 105. V notch versus round notch on a hard, quenched, and tempered steel.

studying the properties at an elevated or lowered temperature, there is much value in testing over a range of temperatures rather than at room temperature alone; so it would be desirable to adopt the Charpy method whenever feasible. Discussion still rages as to

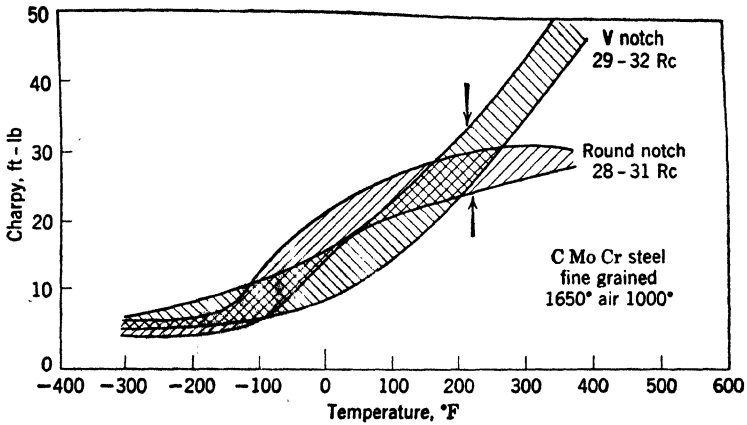


FIG. 106. Another comparison of V and round notches.<sup>34</sup> Steel contains 0.14 C, 0.48 Mn, 0.31 Si, 2041 Cr, 0.53% Mo.

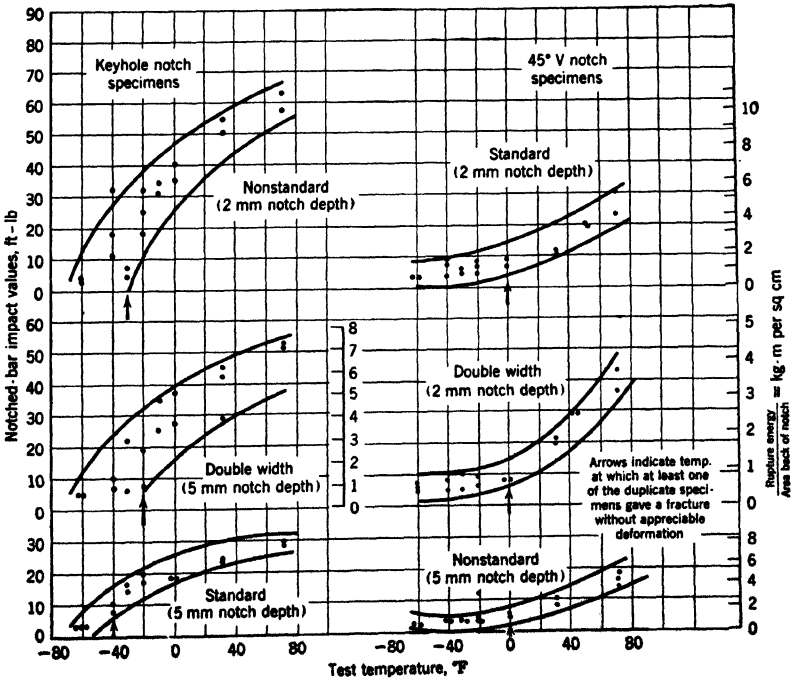


FIG. 107. Low-temperature notched-bar strength values at various temperatures for a normalized medium-carbon-rolled steel of 0.27 C, 0.67 Mn, 0.40% Si, normalized from 1650°.<sup>34</sup>



whether the V or the keyhole notch is preferable. Army Ordnance, and hence its suppliers, appear committed to the V notch. Sieman<sup>6</sup> describes making the V notch in very hard steels.

A comparison of square bars, heat-treated in  $0.4 \times 0.4$ -in. section, with these notches, was carried out<sup>7a</sup> on a variety of steels, the tests being made over a wide range of temperatures (examples are shown

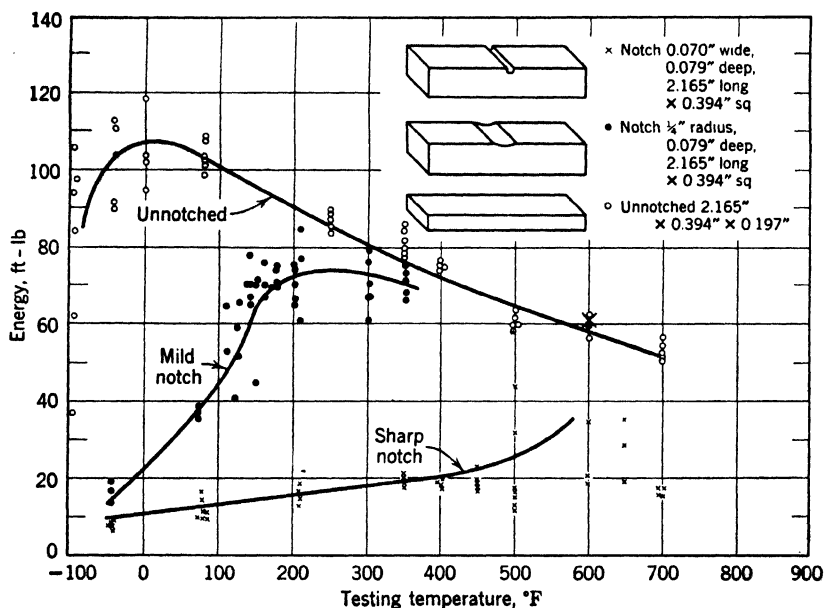


FIG. 108. Effect of shape of piece and of testing temperature on energy absorbed in breaking. (Armiento)

in Figs. 104–107). The temperature at and below which flat brittle patches appeared in the fracture is shown by an arrow on these plots. (The mechanism of fracture which results in the brittle patches is well discussed by Hollomon.<sup>8</sup>) The two bars differ in sharpness of notch and also in area back of the notch, the V-notch bar having the greater area. Figure 107 shows results on “non-standard” bars in which each notch is backed up by the same area. These figures show that the geometry of the bar controls the results and that no reliable interconversion can be made.

The influence of the severity of the notch is shown in work by the Armiento, cited by Grossmann,<sup>9</sup> using a 4150 steel, 0.52 C, 0.85 Mn, 0.24 Si, 1.01 Cr, 0.20% Mo (Fig. 108). The area back of the notches

was the same in the notched bars. The scatter in the results for unnotched bars, from around 105 to around 35 ft-lb is noteworthy. The notches are not standard. By using the sharper notch of Fig. 108, four heats of 4150, agreeing quite closely in chemical composition, were quenched and tempered alike, but gave the varying results of Fig. 109, at room temperature.

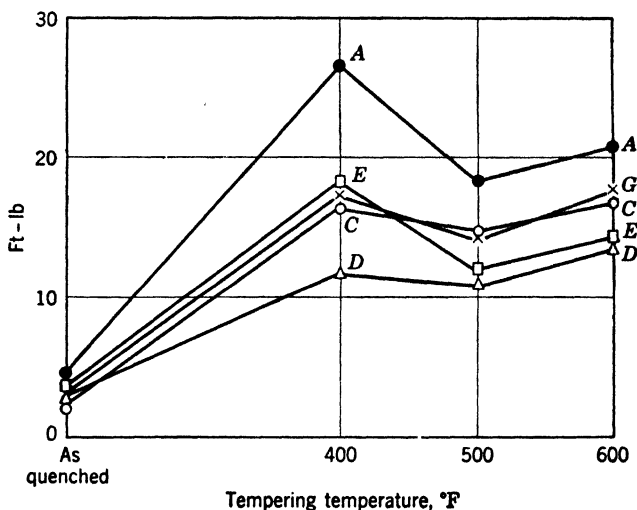
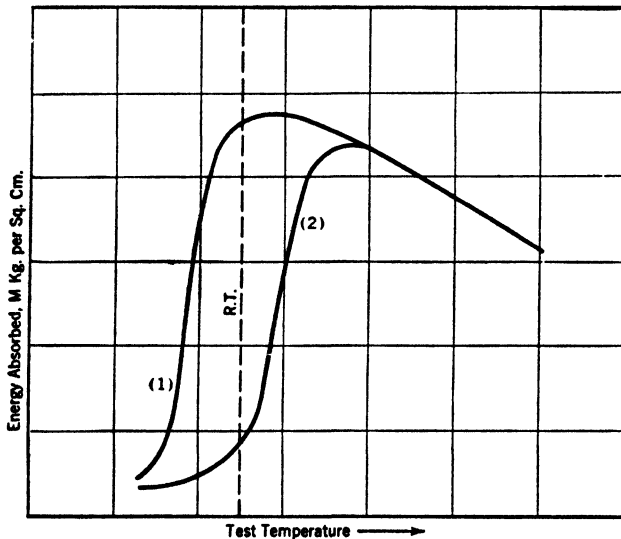


FIG. 109. Notched-bar toughness of five heats of W.D. 4150 steel. (Armiento)

**Influence of Dimensions of Notched Bars.** When either type of test and any one of the common types of notch are used, several peculiarities will be noted when the dimensions of the bar are altered. When the bar width is doubled, the energy absorbed may be much less than doubled and may even decrease. The difference varies with different steels, and one method of evaluation is to find the ratio between narrow and wide bars. This behavior, diagrammatically shown in Fig. 110 from Hoyt,<sup>3</sup> shows clearly that a single impact figure made on a “standard” specimen is not necessarily an adequate criterion for evaluating differences in steels.

For instance, two steels that really differ may behave in a ductile fashion in the test and give the same energy absorbed. This is all the test does tell. It does not tell whether any difference exists in inherent brittleness of the steels, and to find this we would have to increase the severity of the loading conditions until one was ductile and the other not.

**Effect of Temperature on Impact Resistance.** By using the same bar, it is found that a change in testing temperature may radically alter the results, as is shown in Fig. 107 as well as Figs. 109–110. Some commercial steels show this very sharp drop in impact resistance around room temperature, and when this occurs duplicate specimens may give widely different values of energy absorbed. Until



(After Hoyt 8)

FIG. 110. Schematic representation of effect of narrow (1) versus wide (2) bar, or of fine grain (1) versus coarse grain (2) in steel, upon impact results over a range of temperature. With many steels the rapid drop in impact resistance occurs near room temperature. At the higher temperatures the fracture is fibrous; at temperatures below the rapid drop it is crystalline.

it was realized that the impact-temperature curve showed this sudden drop and that erratic test results must follow when the test temperature coincides with it, impact results were considered to be so nonreproducible as to be worthless.

Where no sudden change is occurring, impact tests are reasonably reproducible. At the transition temperature, "duplicate" V-notch Charpy bars have shown 200 and 5 ft-lb, as an extreme case. Compare the unnotched bars of Fig. 108. Note the wide band for the oil-quenched specimens of Fig. 112.<sup>7a</sup> If only a few specimens were tested at  $-40^{\circ}$  and these happened to hug the upper boundary of the band, the fact that the transition temperature has been reached would be missed.

This "transition" is from a wholly ductile tough tearing fracture at the upper temperature to a wholly brittle flat fracture at the lower temperature. The term as used in this connection does not indicate a change of phase, in which sense the word is also used in connection with such changes as are in the equilibrium diagram.

Attention is called to the necessity for showing a band, rather than a line drawn through the average of the individual values, in

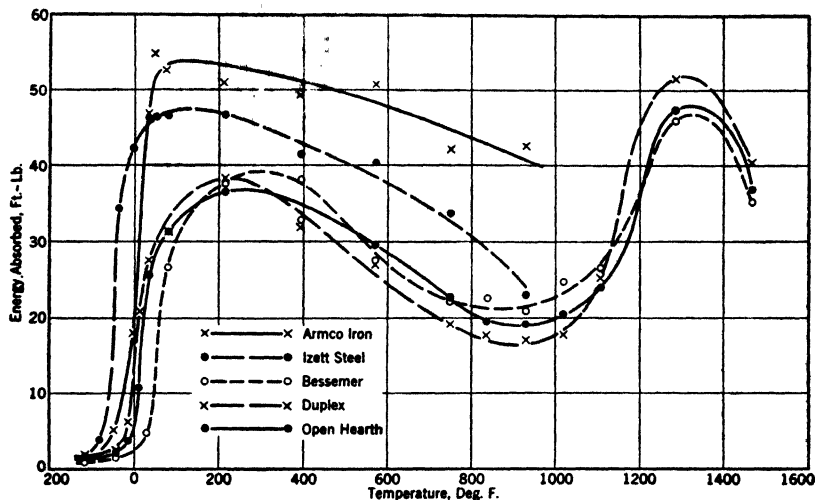


Fig. 111. Charpy impact data (keyhole notch) for low-carbon structural steels. "Izett" is a fine-grained aluminum-treated steel. (Epstein<sup>10</sup>)

plotting a notched-impact versus temperature curve. Compare Figs. 110 and 112. Moreover, it should be noted that failure to make plenty of "duplicate" determinations in the transition region where high, intermediate, and low values can all occur may be very misleading. If only one or a few tests are made and they all happen to be on the high side, the transition temperature may appear to be lower than is actually the case.

The S-shaped curves, such as those shown by Epstein, Fig. 111, are characteristic of the impact properties of practically all ferritic steels. At some range of high temperature, impact resistance will be at a high level and, at some very low temperature, at a very low level. These high and low levels may be higher or lower with different steels, and the band joining them may rise steeply or gradually; but an S type of curve, rather than a straight line, is normal. Although steels quite low in C and high in Ni and a few other types

show exceptions, most ferritic steels become very notch-brittle at liquid-air temperature and almost equally so, though they may be at quite different levels, at room temperature.

No mechanical test is so sensitive to grain size or, more strictly, to grain-boundary conditions, as the impact test. Since properties more troublesome to test for than impact are likewise connected with grain size, the impact test has ability as a convenient mirror of the grain size, quite aside from what it tells about brittleness.

Indeed, it is as a mirror of other difficultly determinable properties or propensities that impact results are perhaps most useful. Temper brittleness, for example, is revealed only by impact.

**Notched Bars as Mirrors of Grain Size.** A high-level of impact resistance at room temperature and a postponement of the sharp loss in impact resistance to decidedly lower temperatures are generally accompanied by fine grain size. This is probably due, not to the grain size itself, but to the accumulation of brittle constituents, such as nonmetallics and carbides, at the boundaries of coarse grains. The effect is loosely spoken of as caused by grain size, since exceptions to this relationship are rare.

**Slack Quenching.** Low-temperature tests often bring out features not very evident in room-temperature tests. In Fig. 112 the room-

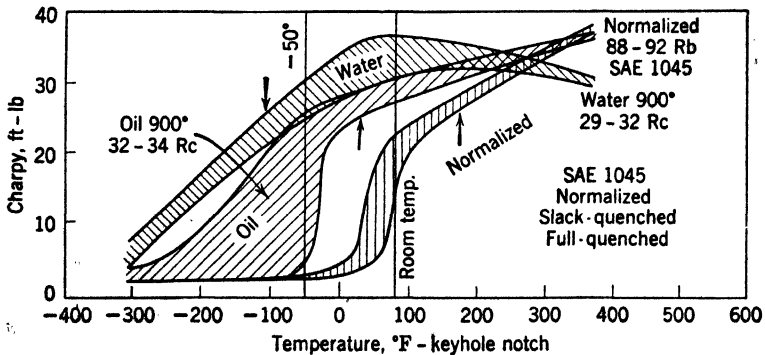


FIG. 112. Effect of slack quenching.<sup>34</sup>

temperature data show the difference between a normalized and a quenched and tempered structure in 1045, but the room-temperature values for water and oil quenching are not widely separated. At  $-50^{\circ}$ , however, the fact is clearly brought out, both by the foot-pound values and the arrows denoting brittle patches in the frac-

ture, that, even in a  $0.4 \times 0.4$ -in. section, oil quenching of 1045 was slack quenching. The low-temperature notched-bar test is a good indicator of slack quenching.

Full martensitic quenching in large sections is made possible by introduction of alloying elements, and in this way alloy content

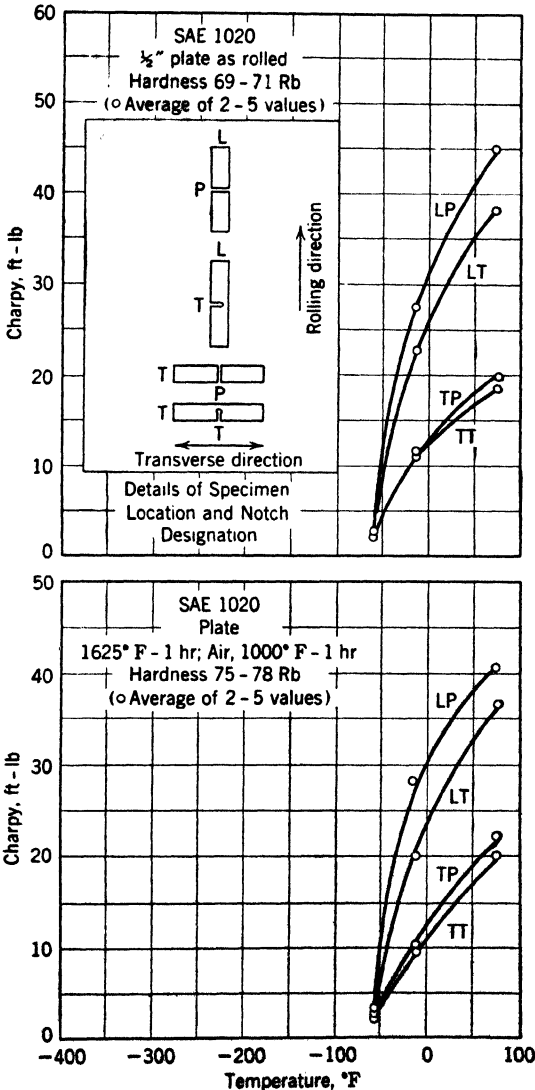


Fig. 113. Directional differences in steel plate.<sup>84</sup>

affects notched-bar behavior, but Hollomon<sup>8</sup> points out that "it is not believed that the alloying elements affect to any considerable extent the impact properties of tempered martensitic steels of the same hardness." Data on many alloy steels are on record.<sup>7</sup>

**Notched Bars as Mirrors of Directional Differences.** Using standard Charpy bars with the V notch, Herres and Jones<sup>11</sup> give, for steels successfully quenched to martensite, then tempered to different hardness levels and tested at  $-40^{\circ}$ , the following expectations for specimens taken in the longitudinal and transverse direction of a plate that has not been cross-rolled.

Brinell Hardness	Ft-Lb at $-40^{\circ}$			
	200	300	400	500
Longitudinal specimens	100+	70	38	12
Transverse specimens	60	35	18	5

Cross-rolled specimens would have intermediate values. One would expect the values in the through-plate direction would be even lower than for transverse specimens.

In 1020 plate, not quenched and tempered, but as rolled or normalized and tempered, Fig. 113 shows a great spread at  $-10^{\circ}$ .

**The Size-Temperature Effect.** The similarity between the effect of increase in section and that of decrease in temperature is brought out by the following tabulation of foot-pounds absorbed for an unkilld 0.24 C 0.46% Mn rolled-steel, longitudinal specimens, V-notch Charpy:

Specimen size	212°	150°	75°	32°	$-60^{\circ}$
Half width	61	60	52	16	4
Standard	68	70	40 *	11 †	4
Double width	64	50	17 ‡	7 †	2

\* Fracture tough.

† Fracture wholly brittle.

‡ Fracture mostly brittle, some deformation.

Another case<sup>7b</sup> of a killed 0.21 C 0.89 Mn 0.22% Si steel, normalized from  $1650^{\circ}$ , gave, when keyhole Charpy specimens were used, in foot-pounds absorbed:

Specimen size	Temperature			
	Room	$-50^{\circ}$	$-100^{\circ}$	$-150^{\circ}$
Standard	55	57	36½	31
Half width	29½	27	22½	22½
Quarter width	12	12	12	12

Whereas the full-size specimen falls off in energy absorbed below  $-50^{\circ}$ , the quarter size remains unaffected at  $-150^{\circ}$ .

Haigh<sup>12</sup> noted the size effect on slow-bend tests of notched bars. Bardgett<sup>13</sup> reported on studies of strain aging, by means of Izod tests on rod and on specimens built up from a pack of thin strips. Embrittlement of the rod by stretching 15% and reheating at  $480^{\circ}$  was evident; the Izod fell from 86 ft-lb as rolled, to 24 ft-lb after straining and aging. The pack of strained and aged thin strips, however, remained tough.

Grossmann<sup>9</sup> cites Bain as remarking that "metals aren't brittle, shapes are brittle." And low temperature helps a shape that has embrittling tendencies, to act brittle. This is not confined to steel. Doan and McDonald<sup>31</sup> emphasize the size and shape effects in Mg- and Al-base alloys.

**Restraint on Deformation.** Metal stressed beyond its yield strength must have space to move into if it is to develop plastic deformation and thus relieve the stress. When the distance to a free edge is small, the metal can upset clear through the section. If the distance is too great, there is mechanical restraint; the center metal cannot push away the outer metal. A notched-bar specimen that breaks with a tough fracture around the edges, but has a brittle undeformed patch in the center, shows this. Indeed, the outline of a fractured bar tells as much as does the fracture itself.

That it is mechanical restraint upon deformation that is concerned, that is, that the notch merely localizes the stress and thus accentuates the need for the metal to flow somewhere and thus distribute the stress, is shown by the behavior of wide widths of unnotched pieces in bending. As Gerard<sup>38</sup> points out, the radius over which a narrow strip will bend without cracking is smaller than that required by a wide strip, because of the added restraint at the center of the large width. The crack starts at that point rather than at the edge.

**Work Hardening.** Comparing the values at  $-150^{\circ}$  in the last tabulation forcibly brings out the virtue of decreasing the restraint. A huge crane hook, made in one piece, is likely to fail in brittle fashion; hence such hooks are made up of many laminations of relatively thin plate, to produce a hook of the same static strength as a one-piece hook, but without its brittleness. The steel in the one-piece hook and the laminated hook could be from the same heat and have the same treatment, and differ only in their embrittling



tendencies, here due purely and solely to geometry. A standard notched bar cut from the one-piece hook and from a lamination of the built-up hook, would show identical notched-bar behavior.

That is not to say that different steels would not behave differently in the one-piece hook. The mechanism of the hook failure is the work hardening of its surface by the chain resting in the hook, the development of a crack in the surface after the work hardening has progressed too far, and the propagation of the crack through

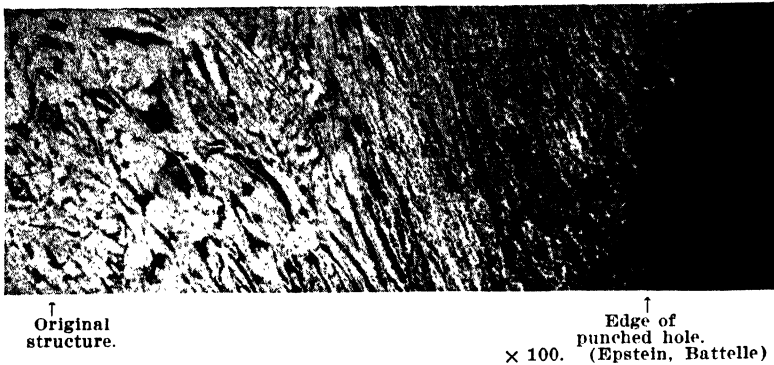


FIG. 114. Distorted structure around cold-punched hole in structural-steel angle  $\frac{1}{2}$  in. thick. This brittle distorted structure extends for about 0.02 in. back of the periphery of the hole.

the body of the hook. Different steels differ in response to work hardening and in resistance to crack propagation. But the conventional impact test, made on material with no cold-worked surface, will *not* reveal the difference. This is directly proved by data from Epstein<sup>10</sup> who studied  $\frac{1}{2}$ -in. angles with holes punched in them. The punching work-hardened the circumference of the hole (see Fig. 114). When the angle is bent to fracture, failure starts through the hole, and the fracture may be either tough or brittle.

This behavior of the actual angle compared with the Charpy impact values of "standard" specimens showed the results given in Table 7. It is the restraint on deformation that counts. The velocity of impact in the conventional methods of notched-bar testing has negligible effect, if any, over that of ordinary slow bending. The transition from tough to brittle behavior is brought out by slow-bend tests as well as by impact tests.<sup>82</sup> The different behavior in Epstein's tests was due to the different steels responding differently to restraint.

TABLE 7

<i>Steel</i>	<i>Fracture in Angle-Bend Test</i>	<i>Charpy, Ft-Lb</i>
<i>A</i>	Brittle	27
<i>B</i>	Ductile	27
<i>C</i>	Brittle	31½
<i>D</i>	Ductile	25½
<i>E</i>	Ductile	31½

**No Relation to Notched Fatigue.** The Charpy bars gave no criterion whatever of actual behavior of full-size specimens in this case. Similarly, the notched-bar impact figures are no indication of the behavior of a steel in notched fatigue. Inglis<sup>14</sup> showed this nicely. A certain Ni-Cr steel part was used for several years at 750°C. This combination of time and temperature embrittled the steel, its notched impact having fallen from the original 38 ft-lb to 4. But endurance tests on threaded (that is, notched) specimens from a bar of the same steel that had been stored at room temperature and from the embrittled piece gave exactly the same endurance limit, 23,500 psi.

**Notched-Bar Tests as Substitute for Ballistic Tests of Armor.** Armies pay a great deal of attention to the type of fracture of armor, especially to the amount of shiny crystalline facets appearing in the fracture of a notched-bar impact specimen broken at low temperature, and an excessive amount is considered a cause for rejection. This is an unfortunate situation from the point of view of inspection, since it introduces a large factor of personal judgment, and actual ballistic tests rather often fail to agree with the prediction made from the fracture, though there is an undeniable general trend toward correlation. Ballistic tests, as ordinarily made, are not so reproducible as they should be, and it is to be hoped that they can be simplified and made more precise so that direct tests can be made with more assurance.

Correlation between a direct and a substitute test needs to be proved before the substitute can be relied on, and the correlation in the armor case seems insufficiently established.

Hoyt<sup>3</sup> remarks, "I have personally observed that a steel which ran 8 ft-lb in the Charpy test was perfectly ductile to the penetration of a bullet."

The difference in velocity of impact of a projectile and of that in the conventional notched-bar test is great, and the steel's behavior

should be far more certainly delineated by a test under service conditions than on a little test bar. It is possible to carry out an Izod type of test by shooting bullets at different known velocities at the top of an impact specimen, instead of hitting it with a tup, and thus to determine the velocity at which fracture occurs, and to note the resulting fracture. If specimens of full thickness were used, this modification might show more satisfactory correlation.

Probably the greatest justification for the use of the standard notched-bar test on armor is the evidence it provides as to full or slack quenching, since the latter is far less desirable in armor.

**Notched-Bar Tests as Alleged Indicators of Steel "Quality."** How much value the impact test has in evaluating materials for service, from the point of view of selecting one material instead of another, is a moot point. It has been said that landing gears for airplanes stand up in service better when made from materials that have high impact resistance, and rail failures in cold weather, though very infrequent, are thought to be related to the very low impact resistance shown by much rail steel in the range of  $-20^{\circ}$ . Nevertheless, when proper fillets are used and sharp notches are generally avoided in design, materials of low impact resistance may serve quite as well as very tough ones.

Riegel and Vaughn<sup>15</sup> consider that notched-bar tests, especially those on double-width bars, combined with engineering experience in their interpretation, are an aid in differentiation between satisfactory and unsatisfactory steels for heavily stressed gears in tractors, when chemical analysis and static properties failed to distinguish between them.

The "quality" roughly measured by the notched-bar test is behavior as to plastic flow under triaxial stresses, stresses acting in three directions. The only simple way to introduce such stresses is by a notch, but stresses so introduced are not readily measured. Plastic deformation is not greatly restrained at the terminals of the notch, the displaced metal can move outwardly, but at the middle of the notch there is no place for displaced metal to move into, and flow is mechanically restrained. Similarly, a lowering of the temperature strengthens and stiffens the metal and likewise exercises restraint upon flow. The intensity of the triaxial stresses is dependent on the shape of the notch. Hence, no single size of specimen nor shape of notch can be chosen to reproduce general service conditions. Two steels may behave equally tough in a standard room-temperature test, but very differently under slightly increased re-

straint. Thus, impact values in themselves are of no direct value in design. A material of 50 ft-lb impact is not ten times as good as one of 5 ft-lb. The two may be exactly as good, or the low one may be of no use at all.

When impact values on a material, so heat-treated that it gives an average of 40 ft-lb rise to 45 or fall to 35, it means little. But when the values fall to 10 or 15, it means that something has occurred that is not according to schedule and we had better look into it. Similarly, if a material that has been in long-time, high-temperature service is removed, tested at room temperature, and shows no change in tensile properties, but does show its impact to be halved or quartered, it means that the material is not wholly stable and that we had better find out what is going on. Thus, the impact test finds its chief use in telling the metallurgist something, at least, about variations that he might not otherwise suspect and serves as a signpost to indicate the necessity for further study of these variations.

But he should be careful not to read into the results more than they really signify. It is especially to be noted that notched-bar impact figures are no indication of the behavior of a steel in notched fatigue.

It must also be remembered that there are large heat-to-heat variations in the notched-bar behavior of otherwise "identical" steels, and that many duplicate tests are needed in the transition range.

Another important point is that the figure for foot-pounds absorbed tells nothing about the type of fracture. Pellini and Queneau<sup>35</sup> cite cases where the fracture of a V-notch Charpy specimen was entirely granular, that is, of the brittle type, with 30 ft-lb absorbed, and where the fracture was mostly granular with 50 ft-lb absorbed. On the other hand, they found fractures that were mostly fibrous and tough at 30 ft-lb absorbed.

**Other Toughness Tests.** The single-blow-impact or the slow-bend test of notched bars is not the only way of evaluating toughness. Impact tests are sometimes made on unnotched bars. For example, rail steel subjected to different heat treatments has been tested by impact on unnotched bars in order to secure a higher level of values and make results more discriminatory.<sup>16</sup> Unnotched bars are used, for the same reason, in testing plastics and zinc-base die-casting alloys, but in general such tests are not thought to indicate behavior under multiaxial stresses as definitely as do tests on notched bars. Tensile-impact tests,<sup>17</sup> in which an unnotched tensile bar is rapidly

broken by shock loading, show differences among steels, but data so obtained are too scanty to allow evaluating the method.

In the study of tool steel, and of toughness on very hard steels (say Rockwell "C" 60 and up, that is, where the Charpy and Izod figures are so low that the tests are not discriminatory), a torsion impact test on an unnotched bar has been devised, but its meaning and interpretation are not yet certain.<sup>18</sup>

Gray<sup>19</sup> shows a plot comparing various ways of evaluating "toughness" in a 1.1% C tool steel at a range of high hardnesses. The Izod curve rises as the hardness decreases, but the curve is wavy, not smooth. The static-bend test indicates maximum toughness at about 650 Vickers, whereas the torsion impact curve shows a deep dip at that hardness, with a high maximum at 775 Vickers. These differences reflect differences in test conditions, proving that toughness is relative to the conditions of service, rather than an absolute property of the steel. It is not clear, however, just what service conditions are approximated by the different types of tests. Gray cites a case where the life of a die, of high-C high-Cr steel, was improved when the hardness was that at which the static-bend test indicated maximum toughness, though there is no obvious causal relation since the die is not subject to plastic bending in service.

**Repeated Impact.** Tests in which a round bar is grooved with a notch, which may be square, as in the Stanton test, or rounded, as in Knowlton's<sup>4</sup> tests, Fig. 101, have been used. The bar is supported at two points with the notch midway in the span, and a hammer is dropped on the bar at the notch. The bar is turned over 180° between each blow, or, in some machines, through a smaller arc between blows. Depending on the intensity of the blow and the properties of the material, fracture will ensue at some moderately reproducible number of blows. The test is little used since it is generally thought that, if fracture occurs in relatively few blows, materials are rated about as by the single-blow notched-bar test; and, if it occurs after many thousands of blows, they are rated much as by the notched fatigue test. No agreement is found as to standard conditions for making the test or as to interpretation of results obtained under such conditions that fracture occurs between, say, 1000 and 100,000 blows.

The test was formerly used in the study of notch sensitivity of steels for automotive use and had some strong adherents. Latterly, it appears to have fallen into disrepute.

**Wedge-Curl Test.** Malleable cast iron is said<sup>20</sup> to be so ductile that it seldom or never fails in compression or transverse loading, since it flows indefinitely, or deforms so much that the part is useless, before it fractures; yet malleable iron will nevertheless normally

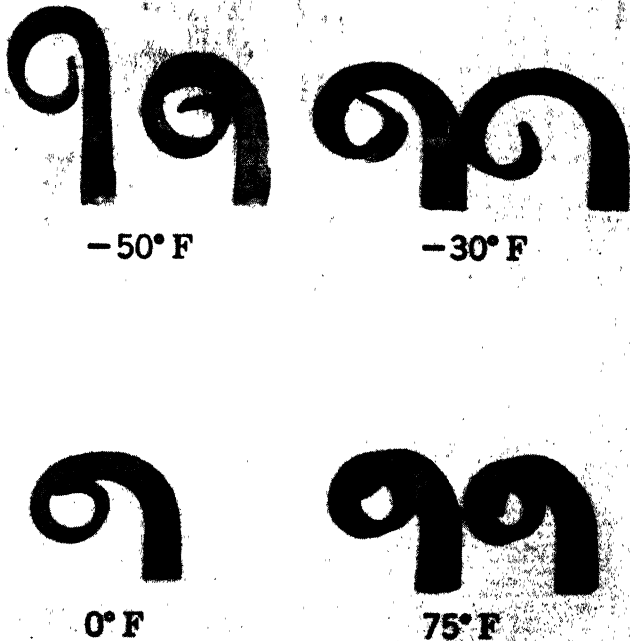


Fig. 115. Toughness tests of malleable iron.

	-50°F	-30°F	0°F	75°F
Single-blow notched Charpy value, ft-lb	3.5-4	5	5.5-6	8-8.2
No. of 70 ft-lb blows on wedge	14-22	22-24	27	30-30

run only about 7 to 12 ft-lb Izod or Charpy. In appraising "galvanizing embrittlement" of malleable iron a figure of 5 Izod or 7 Charpy is considered to show a tough condition, not a brittle one. Malleable is employed for shock-resisting service in railway uses, in road machinery, agricultural implements, and the like. Even when the specifications for elongation in the tensile test called for only 10%, as they did in 1924, instead of the 18% now required

in the high-grade malleable, the material was considered tough and suitable for shock-resisting uses. Frye<sup>21</sup> illustrates its use in ordnance equipment.

Malleable producers evaluate their product by a wedge-curl drop test. A 21-lb tup is repeatedly dropped  $3\frac{1}{3}$  ft (thus giving 70 ft-lb) on a wedge 1 in. wide,  $\frac{1}{2}$  in. thick at the butt, and  $\frac{1}{16}$  in. thick at the tip. The test is stopped after 30 blows, or earlier if a crack appears. Comparison of the results with the standard Charpy keyhole notched bar and with this test on one lot of standard malleable, over a range of temperatures, is made<sup>22</sup> in Fig. 115.

This shows that the notched-bar classification of malleable as "brittle" is not necessarily an exact one.

**Brittle Failures.** A few disastrous brittle failures have occurred. An outdoor hydrogen storage tank at Schenectady blew up<sup>23</sup> on a cold winter day, one side being in bright sun, the other in the shade. A tank for storage of liquefied natural gas at very low temperature blew up at Cleveland,<sup>24</sup> and several welded ships cracked<sup>25</sup> at ordinary temperatures.

Nothing appears to have been published on the steel in the Schenectady tank or its low-temperature properties.

The Cleveland tank is reported<sup>26</sup> to have been constructed with an outer shell of standard C steel, an inner shell of normalized fine-grained 0.10 C  $3\frac{1}{2}\%$  Ni steel, showing 20 ft-lb Charpy at  $-260^\circ$ , and welded with austenitic 25 Cr 20 Ni rod. A crack developed the first time the cold liquid gas was admitted to the tank in June 1943. The crack was repaired (thereafter the tank was more uniformly precooled before being filled). This indicates that thermal stress was considerable.

The ships were of unkilld steel. Their failures started at locations, such as hatch corners, where high stress concentration obviously existed, intensified by roughness of the weld surface. The ship failures started with short tearing cracks of the ductile type, soon changing to a brittle type, in which there was no thinning down of the cracked edges by plastic flow. The brittle failure progressed at something like a mile a second; final failure was very sudden. The early welded ships had no riveted joints, which could slip to relieve stress, or at which the advancing crack could stop, but instead were welded into a unitary structure. Relatively few welded ships failed in this fashion; most gave good service, and so the conditions leading to failure were border line.

Exhaustive experiments were carried out, but not yet reported in full, to study these failures. In attempting to duplicate this type of failure in the laboratory, it was consistently found that small structures made of thin-walled tubes or thin sheet behaved tough when those with heavy sections behaved brittle, though the small structures were scale models of the large ones. Killed steel, with moderately better low-temperature Charpy values than those of the un-killed ship steel, did not always behave appreciably differently in the tests than did the un-killed steel, indicating that geometrical restraint had a large influence.

The failures of notched plates start at the base of the notch, midway between the faces of the plate where restraint is at a maximum. The transition temperature varies, with the same steel, according to whether it is determined on Charpy bars, notched tensile specimens, or on notched full-thickness sheets, and in the sheets of the same thickness, according to the width of the sheet. The transition is shown by the change in type of fracture. The standard Charpy value is not a dependable indication of the behavior of a wide plate.

The adding of a "doubler" at a location of stress concentration, in the expectation of supplying more "brute strength," often makes matters worse, since it adds to geometrical restraint.

Hollomon<sup>36</sup> has discussed the ship-failure problem in detailed relation to the fundamentals involved. He points out that a given steel might show a transition temperature of  $-40^{\circ}$  in a V-notch Charpy test; yet, with a welding crack present, plus existing geometrical restraint, the same rate of loading as in the Charpy test will cause propagation of brittle fracture in the actual ship at  $70^{\circ}$ .

Luther and coworkers<sup>37, 38</sup> developed a "high-constraint nick-bend test," compared its indications with those of the V-notch Charpy test, and found that the indicated transition temperatures were not the same, and on a given series of steels, did not always place the steels in the same order. Another test was the tensile test of a slotted specimen designed to produce constraint. With this they produced brittle tensile fractures at room temperature in longitudinal specimens of 1-in.-thick plate, although in the conventional tensile test the steel had 40% elongation. In the case of the same steel, in the same type of slotted specimen, but in plate only  $\frac{1}{4}$  in. thick, longitudinal specimens had to be cooled below  $-95^{\circ}$  to induce brittle fracture. However, transverse specimens, even of the  $\frac{1}{4}$ -in. plate gave brittle room-temperature fractures.



Anderson and Waggoner<sup>39</sup> also present evidence on the effect of geometrical restraint.

All these extensive tests correlate completely with the early work of Epstein<sup>10</sup> on angles of varying thickness.

The fallacy of expecting a tiny Charpy bar to evaluate a steel in any larger size is difficult to overcome. It happens that the transition temperature for most commercial plates is so close to room temperature that, although the Charpy test shows tough behavior even at  $-40^{\circ}$ , the conclusion that, since a thick plate is not expected to be subjected to so low a temperature, the plate will behave tough in service is actually entirely unwarranted. Overlooking directional properties is likewise unjustifiable.

The important thing is to design in order to avoid notches in thick wide plate; or, if notches are inevitable, to design for the use of thin plate. Some designers have done this. Multilayer pressure vessels, tested to destruction, are stated<sup>30, 40</sup> to fail by a series of tears, whereas those of solid thick-walled construction fail by shattering. Thum<sup>27</sup> pointed out long ago on the failures of pressure vessels: "Thin walled vessels develop shear-type fractures. Thick walled vessels develop separation-type fractures, whereas vessels of equivalent strength but made of nested, closely-fitting units again develop shear-type fractures."

No external impact was needed to produce brittle failures in any of the three cases cited here. The hydrogen and liquefied-gas-holder failures may have resulted from expansion and contraction stresses due to temperature differences in different parts of a rigid structure, designed without freedom for motion, in contrast to the S construction used in high-pressure steam piping to allow movement and hold down stress. Boardman<sup>28</sup> has suggested one method for so doing. Jasper<sup>29</sup> discusses stresses due to temperature differences in pressure vessels operated at high temperatures and emphasizes that the thinner the vessel wall, the less the stress differential. The same principle would hold for a temperature differential at any range of temperature.

Although it is obviously conservative to use material as resistant to the low-temperature notched-bar-impact test, as can be afforded (selecting fine-grained killed steel, quenching and tempering when feasible, even in some cases resorting to austenitic steels), yet the Cleveland case shows that use of material which would ordinarily be considered tough, on the evidence of conventional small notched-bar tests, is not necessarily sufficient to avoid failure. Design to

minimize stress concentration and geometrical restraint is the pressing need.

No conventional impact figure is usable in design, but the qualitative evidence produced by impact tests of large versus small specimens gives a hint as to a design that will "divide and conquer."

To be of value for design purposes, an impact toughness test needs to be made on actual pieces having the exact size and shape to be used in service, that is, the true stress concentration, and to be carried out over the whole range of service temperatures.

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## CHAPTER 8

### SPECIAL TESTS

In view of the many different attributes which a designer may need to utilize, such as those listed by Boulton (p. 110), and the limitations of the conventional tests, there is need for more specialized tests, particularly for those aimed to duplicate or approach the actual conditions of service. The research investigator constantly applies highly specialized tests of other types in the examination of the principles underlying metallurgical phenomena.

The heat treater is interested in those special tests which point the way to modifications in heat treatment which better fit the product for its intended use, or to more economical methods of treatment. Likewise, tests which prove to the designer that failure of a part is not to be cured by shifting from one steel composition to another or by heat treating the steel a bit differently, but only by designing to avoid unnecessary stress concentration, can relieve the heat treater of much grief.

Outstanding among the attributes not revealed by the conventional tests, but desired for many production purposes, are formability, weldability, and machinability.

#### FORMABILITY

Formability in the cold is markedly affected by the structure of steel as produced by differences in annealing or normalizing. Cold formability of the degree demanded in many fabrication processes is not obtainable at high-strength levels and hence is seldom required in quenched and tempered steels.

The directional properties discussed on p. 118 come into most cold-forming problems and necessitate that any test be applied across as well as with the grain.

The final test of formability is the behavior in the shop, but prior evaluation of forming properties by some applicable test would save much shop scrap. Selecting a simple but still applicable test is not

easy, for the amount of plastic deformation a steel will stand before it hardens so much as to permit no further deformation without an intervening process anneal (see Chapter 9) varies with the type of deformation—whether it is upset, stretched, etc.

Many forming processes involve a variety of types of plastic deformation, and a test involving only one type may fail to reflect behavior in an actual process.

In the case of sheet, bulging or cupping tests, such as the Ericksen and Olsen, stretch the metal to failure. Actual deep drawing into cups, on a press, of disks of varying diameters and noting what diameter just fails to draw without rupture probably approach some operating conditions more closely than the bulge test. This technique was originally used on zinc<sup>1</sup> and has been applied to steel.<sup>2</sup>

## TRUE STRESS-STRAIN TESTS

Fundamental behavior during plastic flow is made more evident than in the uncorrected stress-strain diagram, by measuring elon-

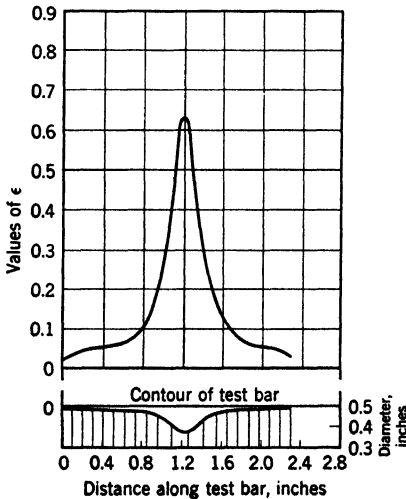


FIG. 116a. Strain and reduction of area along fractured test specimen of cold-worked 1112 steel (as received).

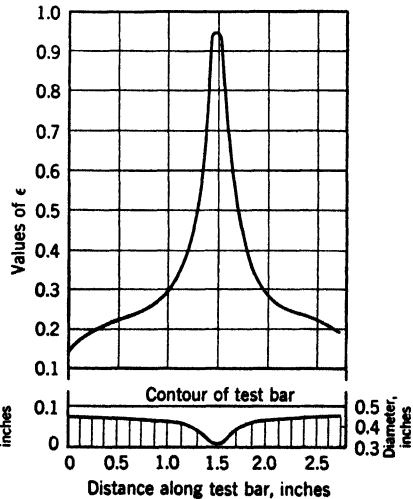


FIG. 116b. Strain and reduction of area along fractured test specimen of annealed 1112 steel.

gation and contraction throughout the taking of a stress-strain diagram, and plotting true stress against true strain. MacGregor<sup>3</sup> shows Fig. 116 for moderately cold-worked and annealed 1112 determined on round bars.

Cold working reduces the ability for general elongation, but much ability for localized deformation remains. This detailed examination is more revealing in a study of cold formability than the usual "ductility" figures. Gensamer's<sup>4</sup> discussion of true stress and strain is pertinent.

MacGregor also described<sup>5</sup> a "two-load" method of approximating the true stress-strain curve, by the use of a tapered specimen, without having to follow the contraction during the test.

The true stress-strain diagram is being extensively utilized in the study of the deep-drawing properties of sheet. From such studies it has become evident that elongations measured on the standard 2-in. or 8-in.-long-by-1-in.-wide test piece used for sheets are not sufficiently informative, since some types of cold deformation depend on the ability of the material for very localized deformation, for example, that occurring in  $\frac{1}{100}$  in. Contraction is not easily measurable on flat stock; hence test pieces of forms more suitable for measurement of general and of local elongation are used to give the type of information gleaned from contraction values obtained with round bars. The heat treater may need to utilize true stress-strain diagrams and special specimens in evaluating the effect of annealing treatments on cold-worked materials.

**Work Sensitivity.** The rate at which a steel work-hardens under cold deformation has a bearing on its cold formability, its machinability, and its resistance to some types of wear. The shape of the true stress-strain curve beyond the yield stress tells something about this property, and special, but as yet unstandardized, tests have been devised to evaluate it.<sup>6, 7</sup>

**Bend Tests.** The ability of a steel to be formed by cold bending is an important property to the fabricator. The bend test measures cold formability quite directly.<sup>8-12</sup> Expensive equipment is not required in the test as usually carried out; a flat test specimen is, for very thin material, bent over flat upon itself, or, for thicker stock, over a pin whose diameter is specified. The diameter is usually stated as some multiple of the thickness of the specimen, the multiple increasing as the severity of the test is decreased. The bent specimen is examined for cracks.

The test may be made more quantitative by determining the local elongation on the outside of the bend, a method used in the study of welds. For thin stock, repeated reversed bending may be used as a

criterion of formability and toughness. Wrap tests on wires are similar direct tests for the ability of a wire to stand the type of deformation it may receive in service.

#### WELDABILITY

Welding is so important a method of fabrication that the suitability of steels for welding bulks large in their selection. Welding is very seldom used on heat-treated steels, since the welding heat alters the structure that has been secured by heat treating. Heat treatment would have to be applied after the structure has been assembled, and most welded structures are too large to be conveniently heat-treated. Even a mere stress-relief anneal on large assemblies involves expense and difficulty.

The relation of heat treating to welding is ordinarily considered to lie in the selection of steels for welding that are *not* too readily heat-treatable, that is, those that have little propensity toward air hardening on cooling from the welding heat, so that the welds and the stock adjacent thereto are not embrittled.

Actually this may not be so vital as has been thought, for a *thin* layer of brittle but sound material, backed on each side by tough material, may not make the assembly brittle. The more important thing to avoid may be the production of hidden cracks.

There is no general agreement as to what is an adequate or applicable test for weldability.

Every sort of welding procedure automatically introduces heating and cooling cycles that produce very localized structural changes at and near the weld, of quite the same nature as are produced by everyday heat treatment of more massive sections; so heat treatment and welding are concerned with the same fundamental principles. This will become more evident in later discussion (Vol. II).

#### MACHINABILITY

Since practically every piece of steel that is heat-treated is machined before or after treatment, or both, the ability of the steel to be readily and cheaply machined is often of paramount importance; and steel is often chosen or rejected because it does or does not fit the requirements as to machinability. A quantitative measure of



machinability would be a boon, but seldom has anything better than a rough qualitative evaluation of this property been obtained. In different types of machining, such as turning, drilling, reaming, and milling, different steels and different structures may stand in different orders. The cutting tool, in its own composition and structure, in its geometric form, and in the speed and feed at which it meets the work, affects the machinability of the work. Even in grinding, the particular abrasive used, the size and shape of the particles, and the nature of the bond have to be chosen for the particular job. Moreover, the chief criterion of machinability may be the speed with which the steel can be machined, the life of the cutting tool, or the quality of finish; and all these may enter into the over-all cost, which is the ultimate criterion.

Wide differences in behavior on machining can be detected and described, but small differences, such as those between different grades of "free-machining" steels, are very difficult to evaluate outside the machine shop itself, and even then different shops may rate a series of steels in quite different order.

In spite of the difficulty of exact appraisal of machinability, gross differences may be appraised by simple tests. "Sawability," determined by noting the time it takes to saw through a bar of steel, chosen as the standard for comparison, and the time to saw through the same-sized bar of the steel being tested under the same pressure on a power hack saw; and "drillability," similarly determined by the time it takes to drill a given depth into a comparison standard and into the steel being tested, using a power drill press and the same pressure, are useful determinations. Much care has to be taken to ensure that the saws or drills used have the same cutting power in a series of tests, so that taking a cut on the standard steel before making the test and again after the test, to make sure that the tool was normal at the start and only normally dulled during the test, is an important precaution. Uniformity of the comparison steel chosen is obviously necessary.

It is possible that well-controlled turning tests, now in process of evolution, will ultimately materially increase the ability to measure machinability from the standpoints that the machine shop is interested in.

As befits the importance of the problem, there is a great mass of literature on machining and machinability. The heat treater is primarily concerned with a few specific questions which are discussed in Vol. II.

## WEAR RESISTANCE

Resistance to abrasion by nonmetallic materials and to attrition by one metallic part moving upon another is required in many services, but wear resistance, like corrosion resistance, is a function not only of the steel itself but equally of the conditions of service.

There is no inherent and measurable wear resistance of steel as an innate property, and correct promises of performance in service can only be had in the laboratory by devising a specific wear test reflecting the particular service condition to be met. An ASTM symposium<sup>13</sup> has dealt with wear and wear testing. Examples of some specialized equipment to meet special problems have been given.<sup>14</sup>

As a first approximation, wear resistance tends to rise with hardness, as long as this does not involve extreme brittleness and chipping or spalling of the surface. The presence of finely divided and uniformly distributed particles of free carbides in a softer matrix adds resistance to most types of wear, especially when the carbides are of a particularly hard nature, as in Cr steels. A sufficient amount of C to provide for the presence of such particles is therefore found in most tool and die steels.

When the surface pressure is great enough to produce a work-hardened skin, such as is characteristic of austenitic steels, particularly of the high-Mn steels (Vol. III), good wear resistance is produced in these steels but if the service is not of this type they perform poorly.

Parenthetically, wear may not always be a bad thing. The brittle behavior of a work-hardened and cracked skin has been referred to (p. 166). It would therefore appear that the work hardening of a railroad rail by the rolling load would involve hazard. However, in this case, abrasion is also taking place, some of the deformed metal being progressively removed by this action. Rail steel has low ductility compared with most engineering steels; yet it withstands this local displacement of metal without the development of a sudden failure. The smoothing of the surface by abrasion and by the peening action of the rolling load tends to remove and iron out the deteriorated skin instead of starting a stress-raising crack pointed in a direction to do harm.

In metal-to-metal wear if both mating surfaces are rubbed clean from any intervening film, so that true metallic contact is made,

local welding may occur, even under rather low pressure, resulting in galling or seizure. Grit caught between the surfaces can act as a cleaner, and thus favors seizure. Avoidance of this type of wear depends on providing a stable adherent nonmetallic coating of some sort on the surface, and in avoidance of grit. Service conditions need to be duplicated in studying the galling type of wear.

**Shearing Strength (Torsional Strength).** The shearing strength of steel is of significance in considering the strength of shafts and axles, which are subjected to torsion, and of bolts and rivets, which often are subjected to direct shear. The torsion test is the test best adapted to determine elastic strength in shear; and by the use of hollow specimens of suitable design the ultimate shearing strength of steel can be determined. It is doubtful whether the commercial determination of shearing strength will become common, because for most steels the shearing elastic strength lies between 50 and 60% of the yield strength; hence the yield strength in tension serves fairly well as an index of shearing strength. The ultimate shearing strength of steel usually is somewhat more than 60% of the tensile strength.

**Compressive Strength.** The compressive strength of steel is best determined by testing small cylinders with a height equal to two or three times the diameter. Such cylinders must have their ends machined very carefully to plane and parallel surfaces, must be centered very carefully in the testing machine, and load should be applied very slowly through spherical-seated bearing blocks.

Only for very brittle steels is it possible to determine the ultimate compressive strength. For all ductile steels the specimen flattens out under load, and there is no well-marked fracture. The elastic strength of steel in compression can be determined by using apparatus and methods similar to those used for determining elastic strength in tension. The compression test is not much used as a commercial test for steel, because most available test data indicate that the elastic strength of structural steels in compression is not widely different from the elastic strength in tension; hence the tensile test serves fairly well as an index of compressive strength. However, the compression test may become an important test for structural steel for columns, since this directly determines the property utilized in service.

Structures built up from thin sheet, as is common for certain aircraft parts, are prone to buckle, so that one side of the sheet is in compression while the other side is in tension. If either the com-

pressive or the tensile yield strength is low, plastic deformation will start on the side with the low yield strength. In such design, equality of the yield strengths is a virtue. This virtue exists in quenched and tempered steels and in normalized or annealed steels, but not in cold-rolled steels, as is discussed in Chapter 9.

### NONDESTRUCTIVE TESTS

**Radiography.** A potent reason for an actual structure to have mechanical properties inferior to those predicted from conventional tests, or even from simulated-service tests on a full-sized sample, is the unsuspected presence of subcutaneous flaws, porosity, slag pockets, blowholes, cracks below welds, etc. Radiographic examination by X rays or by gamma rays from radium reveals some types of large flaws. Such examination is now a commonplace, and there is ample literature about it.<sup>15-16</sup>

**Supersonic Examination.** A rather new method, difficult in application and even more difficult in interpretation, seeks to locate flaws by reflection of supersonic waves from such discontinuities.<sup>17-19</sup>

Desch, Sproule, and Dawson<sup>20</sup> discuss supersonic testing of steel, in considerable detail, concluding that it can be made to give certain information about the internal character of masses of steels, notably sponginess and cracks, but much work on equipment and technique would be required before it could be classed as a practical method for acceptance or rejection. Active development work continues.<sup>48</sup>

**Identity Tests.** A new method<sup>21</sup> for checking the identity of one material or specimen with another relies on triboelectricity. This has not had sufficient use to allow appraisal. There has been more experience with a device called the Cyclograph, which compares a standard specimen with another whose identity with or difference from the standard is to be established. Both pieces are brought into a magnetic field and their behavior in response to magnetization is compared through the trace on the screen of an oscilloscope.

The Cyclograph and the other magnetic-analysis methods<sup>22-25</sup> tell only whether the two pieces differ, not *how* they differ. Identity tests will often detect a mix-up in steels.

Radiography and supersonic testing may serve to detect pieces that should be discarded without heat treatment, but have few possibilities for further evaluating the heat-treated product. Identity tests may have further application in showing whether heat treatment has been uniform from piece to piece.

**Magnaflux.** Heat treatment may produce quenching cracks in steel, not detectable by visual examination nor by X rays. Such cracks, grinding cracks, cracks produced in service, slag stringers near the surface, and similar surface imperfections are made visible by magnetizing the steel and applying magnetic powder. The powder is easily removed where no crack or flaw exists, but clings to such a defect. The magnetic-powder technique has been highly developed, and authoritative literature on both the method and its interpretation is available.<sup>26-31</sup>

**Magnetic Testing.** Certain magnetic tests are employed in the evaluation of material for use in permanent magnets, transformers, and other electric equipment, too specialized a matter to discuss here. From the viewpoint of heat-treatment problems, it is sometimes of interest to know the amount of austenite retained after quenching; this can be roughly indicated by magnetic methods. More precise methods are badly needed. Magnetic methods may also show flaws.<sup>47</sup>

**X-Ray Spectrometer, X-Ray Diffraction, and Electron Diffraction.** The X-ray Spectrometer may also be applicable to the determination of retained austenite, but this use is not yet thoroughly developed.

X-ray diffraction is an adjunct to the crystallographer. Evidence as to constitution of phases present in alloys, submicroscopic grain size, the spacing of the atoms in the crystal lattice, and other bits of information which even the electron microscope cannot reveal, are developed through such methods.<sup>32-35</sup> Many of the conclusions as to principles of metallurgy are based on X-ray evidence of such sorts. Milburn<sup>36</sup> cites a practical use in determining the depth to which shot peening alters the surface.

Electron diffraction is similar in nature, but especially useful in the study of surface films.<sup>37</sup> All these methods require special technique and special experience in interpretation of results. The average metallurgist has to take on faith the findings of the physicist specialists in these lines.

**Damping.** The ability of material to dissipate energy by converting it into heat within the material is known as its damping capacity. A steel tuning fork does not dissipate much energy and keeps on vibrating for a long time with only slow diminution in amplitude, whereas a similar fork of cast iron ceases vibrating very soon. Damping capacity varies with structure and heat treatment. It has been studied in relation to fatigue without noteworthy results but may perhaps be a useful tool in the study of the effects of reheating

after cold work and of structural changes resulting from heat treatment.<sup>38-45</sup> Damping is sensitive to structure. Hence, though a very indirect method, it might serve as an identity test.

In special heat-resisting alloys for gas-turbine construction, damping ability at operating temperatures is considered a virtue. Special methods<sup>46</sup> are used for its determination.

**Low-Temperature Properties.** The futility of evaluation of notched-bar behavior on the conventional specimen was brought out in the previous chapter. Low-temperature notched-bar behavior is not predictable from room-temperature tests, and so testing at a range of low temperatures is called for as a part of the proper technique of the conventional test. It is no longer to be classified as a special test.

**High-Temperature Properties.** For permanent resistance to high loads at high temperatures, C steel has extreme limitations; its load-carrying ability is low and it is too readily oxidized. Very large amounts of suitable alloying elements must be used to overcome these deficiencies.

However, much C steel is used at temperatures low enough for oxidation difficulties to be absent. At quite low temperatures the usual observation that fine grain is desirable still holds, but at higher temperatures, including those at which C steel is still usable, the situation reverses and coarse-grained material has the better long-time load-carrying ability. Hence, for such service, heat treatment to coarsen the grain is in order.

The long-time load-carrying ability must be evaluated by "creep" tests, for at elevated temperatures plastic deformation occurs slowly, and time must be given for the deformation to occur. Attempts to evaluate load-carrying ability by attempting to determine the proportional limit in a short-time high-temperature tensile test are fruitless; very high and erroneous results are obtained, nor can they be "corrected" by application of some mathematical factor. Long-time load-carrying ability is only appraisable by very long tests, 1000 hr at least, and with the use of apparatus allowing great precision of measurement and extremely close control of temperature.

Beside the minuteness of the plastic flow to be measured, necessitating a long period during which measurable deformation may accumulate, the question always arises whether the measurement at, say, 1000 hr, allows safe extrapolation to what may be expected at, say, 20,000 hr, or whether the piece is not being heat-treated during the test itself and its structure and properties being changed. Hence

the question of stability at operating temperatures is quite as important as the rate of flow at any given period in the test.

**Stability.** Examination of the unfractured specimen after creep testing, for hardness, tensile strength, and ductility, or, alternately, for impact resistance, often shows that, under the temperature, time, and stress used, some properties have changed. Often, too, a metallographic examination shows a detectable change in structure.

That is, the creep (or the stress-rupture) tests have been made on a material that was in process of change. The early part of the test reports the behavior of the structure then existing, the latter part that of the structure existing after more or less lengthy sojourn under the conditions imposed. The results are, then, a composite. Since the designer wants to know how a given starting material behaves under the imposed conditions, it is this composite he is after. But it often occurs that a suitable preliminary heat treatment that puts the material into a more stable condition at the start of service results in better service, higher load-carrying ability after long use, or reduced tendency toward embrittlement. Hence special studies to evaluate and control stability are often fruitful. These may be time-consuming and difficult, since it is necessary to untangle that which results from those factors combined with the applied stress and whatever deformation results from that stress. If the applied stress is far above the design stress, the resulting deformation may so alter the conditions that they no longer resemble those of service.

#### STRESS-RUPTURE TESTS

Overload tests at high temperature on a series of test bars, the loads being selected so that the bars *break* in anything from a few hours to, say, 1000 hr, evaluate the plastic behavior occurring under these conditions. Attempts are often made to wring predictions out of such data by plotting and extrapolation, as to probable behavior at the lower loads that can be used in design. There are too many exceptions and reversals of behavior to make such extrapolation reliable practice, to say nothing of the lesser ability of stress-rupture tests to evaluate stability, as compared with creep tests.

When large deformation is permissible and the required life is very short, the indications of such a series of tests are some guide to load-carrying ability but one that, even then, needs to be used with caution.

Such tests are chiefly useful for their information on a type of embrittlement to which many high-temperature alloys are prone. Fracture occurring under high load in a few hours may be accompanied by quite high elongation. Fracture occurring under lower load (but still in excess of usable service load) may be accompanied by much lower elongation; in the worst cases the bar may develop many cracks and fail to show appreciable plastic flow, that is—real embrittlement. It is worth while to know whether or not such a tendency is present. However, embrittlement appears generally to result from changes in the metallographic structure that occur but slowly even at the high temperature and are either due to, or accelerated by, the plastic deformation due to overstressing. Many of the alloys that show marked tendency toward embrittlement in overload stress-rupture tests, remain satisfactorily tough for very much longer periods at the loads to which design is limited because no very appreciable deformation can be allowed and have the piece still function.

In early stress-rupture testing no account was taken of the progress of deformation; only the load and the elongation after fracture were recorded. Nowadays the progress of deformation is observed and recorded, just as in a creep test, to give a better picture of what is going on.

Good stress-rupture behavior often changes to poor creep behavior at design loads. Poor stress-rupture behavior is no sure criterion of unfitness, but it serves to put one on guard in respect to the stability of the material.

No designer today would attempt to classify steels for high-temperature service by the conventional room-temperature tensile and impact tests which he would often accept as adequate proof of fitness for other types of service. Yet these other services may depend on properties as little measured by the conventional tests as those of high-temperature service. Since the high-temperature field is one primarily for high-alloy steels, detailed discussion is left for Vol. III.

#### ELUSIVENESS OF UNMEASURABLE PROPERTIES

If all the properties desired in steel could be quantitatively measured and if the user could state definitely just what degree of each property his service demanded, the evaluation of steel and its heat treatment to give it the desired combination of qualities would be easy. A formula could be set up in which each measured property



was weighted as to its importance and a final figure of merit arrived at. Indeed, many "merit indices" have been proposed, to little avail.

In our present stage of ignorance on the testing of steel, the procedure is for the user to measure what properties he can, to observe how steels varying in those properties serve his needs, and to specify as well as he can what minimum properties he thinks he wants, in order to have assurance that the performance of another piece of steel will duplicate satisfactory past performance. In this he is likely to demand, just for security, properties that are not truly essential to his service, and thus to raise the cost through the expense for selection, treatment, and inspection for unnecessary qualities. However, until more quantitative measurement of the necessary, but elusive, qualities is feasible, this is the only avenue open to him. The need for development of specialized simulated-service tests, supplementing or even displacing the conventional tests, and directly aimed at giving the needed information, is abundantly evident. The ASTM has set up a special committee on simulated-service testing.

#### CONVENTIONAL VERSUS DIRECT TESTS

Omission of actual requirements and imposition of unneeded requirements are handicaps to all who have a hand in the steps of production, to the heat treater no less than to the steel maker. Hence the heat treater may well join in the growing demand for simulated-service tests aimed to evaluate *directly* the suitability of a steel or a heat treatment for actual performance. He should recognize, and help the engineer to recognize, that although the conventional tests provide a useful and indispensable preliminary screening, they go no farther. Seldom, indeed, are they crucial tests. Remembering this and that directional and size effects limit the validity of the conclusions likely to be drawn, unless one knows the way the specimen was selected, and always remembering that each heat of steel has its own individuality, we may use the conventional tests as measures of the properties accompanying different structures in steel.

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## SECTION III. HEAT TREATMENT IN THE PEARLITIC DOMAIN, BELOW THE CRITICAL RANGE

### CHAPTER 9

#### COLD-WORKED STEEL AND ITS HEAT TREATMENT. HYDROGEN AND ITS REMOVAL

Properties are what we want; structure is what gives us properties. Fineness of structure tends to give strength and toughness; coarseness tends toward softness with mediocre toughness. Heat treatment controls the fineness of the structure.

Fine structure can also be produced by mechanically breaking down the coarse structure through cold work. This strengthens, but toughness falls rapidly at an early stage of cold working. The deformed fine particles are in a strained and unstable state and want to go over to the more stable state of equiaxed crystals. Heating, usually to  $900^{\circ}$ – $1200^{\circ}$ , allows recrystallization to start, coarsening the structure somewhat, dropping its strength, but restoring some toughness. By control of time and temperature of this heating, termed "*process annealing*," a compromise of properties is effected.

In this process there is no change in the nature of the crystals, only in their geometrical shape and size. The heating is below the "critical temperature," the line *PS* of Fig. 3, at which the crystals do change their nature. Austenite and its transformation products are not involved.

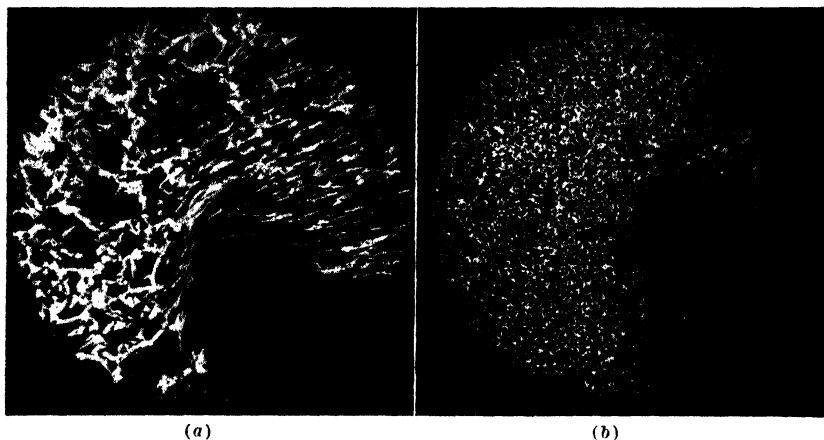
Before process annealing can be considered, we must first note the effects of cold work itself on the properties of steel.

**Methods of Cold Working.** Cold work involves permanent plastic deformation. The yield strength must be passed, for plastic deformation to result. Cold rolling of sheet and cold drawing of rod, shapes, or wire are common methods. Spinning, impact extrusion, etc., are applied to softer metals, less often to steel. Shearing and punching cold-work the metal near the cut, as was shown in Fig. 114.

Even ordinary machining cold-works a thin layer below the cut, as shown in Fig. 117. Hull and Welton<sup>1</sup> find that coarse machining

raises the surface of 1020 steel from 120 Knoop to over 200. Less brutal methods, such as honing or pickling off about 0.02 in. on a ground surface, 0.01 in. on a machined surface, or 0.001 in. after an emery polish are considered to be required to remove the cold-worked layer.

Stamping and deep drawing, as in forming an automobile fender, forming the top of a tin can, or bending the body blank to form the



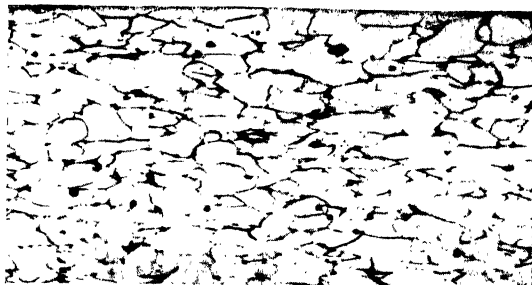
× 235. (Wysor, Bethlehem Steel Co.)

FIG. 117. Cold-worked structure obtained in machining a 0.50% C axle. The cut was interrupted and the tool backed away from the uncompleted cut. The surface of the completed cut and the part ahead of the nose of the tool which will form the chip are badly distorted in the coarse-grained specimen (a), which is in the as-forged condition, 149 Brinell. The surface and chip of (b), which is the same steel in the normalized condition, 163 Brinell, is not so badly distorted, owing to the finer grain and greater hardness.

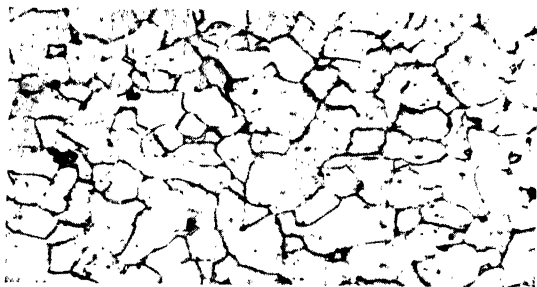
side seam of the can, are familiar examples involving ability for deformation in fabrication.

Such deformation can be very slight or can be carried nearly to the ultimate breaking load and deformation allowed by the stress-strain diagram (Fig. 77).

**Cold-Worked Structure.** The crystal grains of the steel are stretched out or forced out into elongated leaflike plates, and the non-metallic inclusions are broken up and strung out, giving a structure of the type shown in Figs. 118*a* and 119*a*. The further this deformation goes on, the harder and stronger the metal becomes, within limits. In very small sizes, phenomenal strength can be obtained. By suitably cold-drawing piano wire further, and stress relieving, a

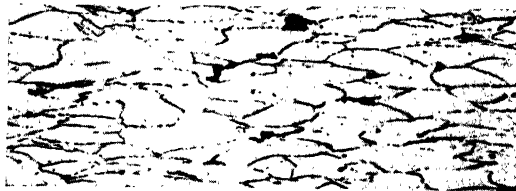


(a) Cold-rolled 40%.  $\times 200$ , 215 Vickers.

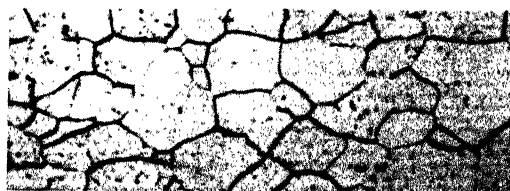


(b) Process-annealed 6 hr, 1275°.  $\times 200$ , 80 Vickers.

FIG. 118. Grain size and shape. Analysis: 0.06 C, 0.31 Mn, 0.02 P, 0.025% S. Aluminum-killed. (Boulger, Battelle)



(a) Cold-rolled 60%.  $\times 500$ .



(b) Process-annealed, 14 hr, 1315°.  $\times 500$ .

FIG. 119. Grain size and shape. Analysis: 0.05 C, 0.38 Mn, 0.025 Si, 0.01 P, 0.015 S, 0.06% Al. (Boulger, Battelle)

tensile of 400,000 psi has been secured. The yield strength is ordinarily raised much more than the tensile strength, though the proportional limit may be low. The usual qualitative inverse relation between strength and toughness is present but in a different degree than that met in material not cold-worked; for, in cold working, the toughness drops very rapidly, becoming very low after relatively small amounts of deformation, thereafter slowly dropping to an even lower figure, so that, with large deformation, not much further ability for enduring deformation would be expected. As it is commonly termed, "the ductility is exhausted." This embrittlement may be put to use. The cold-drawing process is often applied from the desire to produce easier machining, the steel used often being high in S, P, and N for the same purpose. The cold work is thought to help embrittle the steel so that the tool may more easily wedge the chip off by cracking ahead of the tool.

**Ability for Further Deformation.** Some nonferrous materials have almost inexhaustible ductility; gold can be beaten into extremely thin foil, all in the cold, without intermediate anneal, but steel is far from being so amenable.

Cold work does not erase the ability for local deformation so rapidly as it does that for general extension (compare Fig. 116). Even after being cold-rolled to a high percentage deformation, in order to produce very high strength, steel is still capable of more severe bending than one would expect.

In 1010 cold-rolled to 105,000 tensile, 103,000 yield (longitudinal) in 0.018 in.-thick sheet, the elongation over 2 in. is only about 1%; but over a 0.1-in. gage length the local elongation is about 140% and the reduction of area about 40%; the sheet will bend in any direction 90° over a radius twice its thickness.<sup>2</sup>

Many deep-drawing processes, however, require more deformation than steel will stand after it has been cold-worked to such a degree, and so, if the first step in the deep drawing has produced very severe cold work, even locally, some ductility has to be restored so that further deformation can be endured.

**Annealing.** Full restoration requires a "full anneal," but a partial restoration is afforded by a "process anneal," so called because it intervenes between steps in the process. A process anneal can also be used to increase the ductility without unduly decreasing the strength, that is, this is "letting down the temper." Thence, comes the word *tempering*. The noun "temper" originally related to the hardness conferred by a certain degree of cold work, that is, "full-

hard temper" is cold-worked almost all the material will stand, "half-hard temper" is only partially cold-worked. Heating so as to shift "full hard" down toward half hard or full soft then became known as tempering.

The difference between a "full anneal" and a process anneal is that the former destroys the ferrite-pearlite or pearlite-cementite structures, transforming them to the different crystal form, austenite, through heating past the critical temperature; whereas the process anneal, with the temperature held below the critical, produces recrystallization of deformed crystals into undeformed or less deformed ones, without changing their nature (Figs. 118*b*, 119*b*).

When the process annealing of cold-worked sheet or wire is done at the steel mill, instead of being termed "process annealing," it is called "box annealing," because heat travels so slowly through packs of sheets, or coiled strip or wire, that the heating times are excessively long, and heating in the open air would result in ruinous scaling. Hence the sheet packs or the coils are enclosed in a box or bell into which a protective atmosphere is introduced (see Vol. II).

**Internal Stress.** Cold work induces internal stress, by which the material is held in an extended or compressed state, but within the elastic limit. If the elastic limit is exceeded by the stress, the metal flows plastically, and the stress is relieved to whatever extent the metal does flow. The internal stresses are both tensile and compressive, for the two must be in balance, but the location within the section may vary. That is, the outside may be in tension, the inside in compression, or vice versa. Compressive stresses are often desired. Guns, tanks to contain gases at high pressure, etc., may be tightly wound with wire so as to exert compression on the inside, or the guns are expanded by internal hydraulic pressure, so that, when this pressure is relieved, there will be residual compressive stress, so located that, when the gun is fired, the bursting pressure has to overcome the compressive stress, the result being that the effective tensile stress is decreased. Thus the gun is strengthened much as when an outer gun tube is shrunk upon an inner tube or when the gun is tightly wound with wire. This "autofrettage" process is widely used at arsenals, including those of the United States. The subsequent low-temperature treatment is essential in giving the gun the yield strength desired against the internal pressure of the charge.

The method is applied both to forgings and to centrifugal castings. Colonel Dickson<sup>3</sup> has described the process as applied at Watertown Arsenal to centrifugally cast guns. In one case a 0.35 C, 0.60 Mn,



0.30 Mo, 0.10% V steel, after normalizing from 1740° to break up the cast structure, was full-annealed by furnace cooling from 1360°, then water-quenched (twice) from 1360° and drawn at 1290°, after which it showed 96,000 tensile, 77,000 proportional limit, 24% elong., 60% R.A., 35 Charpy on tangential specimens.

It was then cold stretched by enlarging the bore some 40% under internal pressure of 105,000 psi and soaked at 570°, when the properties were:

114,000 tensile, 103,000 proportional limit, 12% elong., 51% R.A., 27 Charpy.

Since metals stand up in fatigue against compressive stress better than against tensile stress, producing residual compressive stress on the surface of a part that must resist repeated tensile stress is a help. Shot peening of parts subject to repeated stress is often highly efficacious in improving fatigue strength, both the increased static strength and the induced compressive stress at the surface being beneficial. Great care must, however, be taken not to over-cold-work and embrittle the surface through the formation of tiny cracks.

Thus it may be desirable to produce and retain a definite distribution or balance of internal stress. However, if the stressed piece is slit, or machined so as not to be symmetrical with its original form, the stresses are redistributed and the piece warps, the lips of the slit move apart or come together. There is a tendency for relief of stress and redistribution of stress by submicroscopic flow, even at room temperature, so that castings that contain residual stress, resulting from the outside cooling faster than the inside, are often allowed to "season" so that the distortion, accompanying relief, and redistribution of stress occur before the casting is finish-machined. Even low-temperature heating facilitates this relief and the warping that accompanies it, so that warping at room temperature thereafter is avoided.

Thus a "stress-relief anneal" may be applied, and at so low a temperature that no visible recrystallization of cold-worked metal ensues, and without markedly altering the strength or ductility.

The normal shrinkage occurring as a welded piece cools produces internal stress. Similarly, cooling after tempering of a quenched steel may introduce some residual stress, and a quenched untempered steel has very high residual stress, as is shown by the propensity quenched steel has toward cracking, which may sometimes be avoided by tempering promptly after quenching.

**Stress-Relief Annealing.** Cold-drawn rod is prone to contain internal stresses, their intensity and distribution depending on the particular technique used in drawing, and so a stress-relief anneal often has to be applied to such products to reduce the tendency to warp, or even to crack. Irregular parts machined from cold-drawn

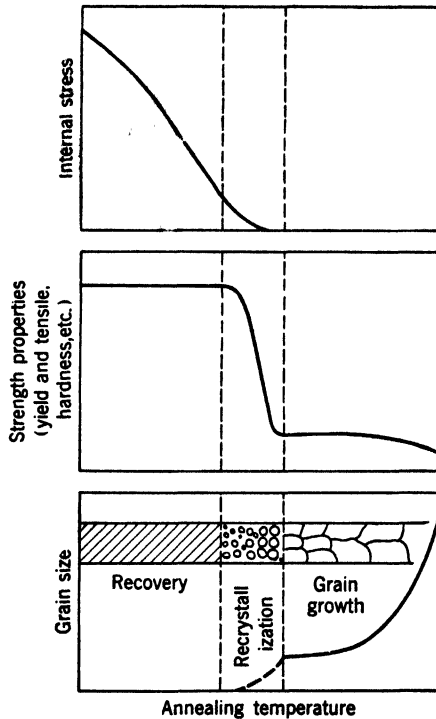


FIG. 120. Diagrammatic representation of the different types of changes in properties produced by annealing a cold-worked metal.

bar stock may warp and change dimensions. A redeeming feature is that many irregular contours may be produced directly by drawing through a die of suitable profile so that finished parts can be produced by just sawing off the desired length. In such a case, when the proper stress distribution has been produced in drawing, the surface stresses thus imparted are retained in the finished part.

Figure 120 from Sachs and Van Horn, whose book<sup>4</sup> should be consulted for a full discussion, shows in diagrammatic form the way internal stress may be relieved by stress-relief anneal below the temperature at which visible recrystallization occurs; it also brings out

the fact that, for *complete* release of internal stress, recrystallization has to occur. The plot is diagrammatic and does not bring out the

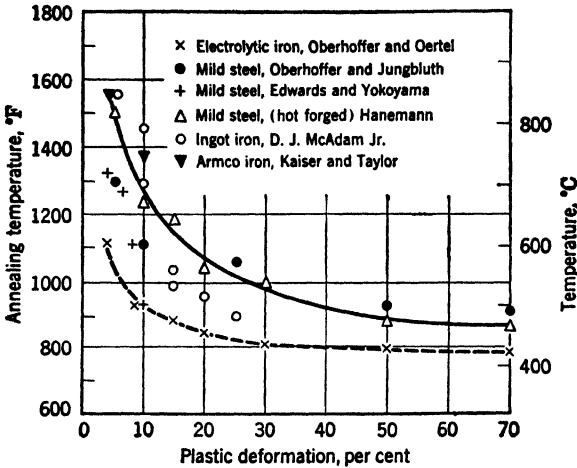


FIG. 121. The recrystallization limit of ingot and low-C steel. (Kaiser and Taylor)

further fact that the more severely the metal has been cold-worked, the lower the temperature at which it will recrystallize and soften. This is brought out in Fig. 121, shown by Sachs and Van Horn, from

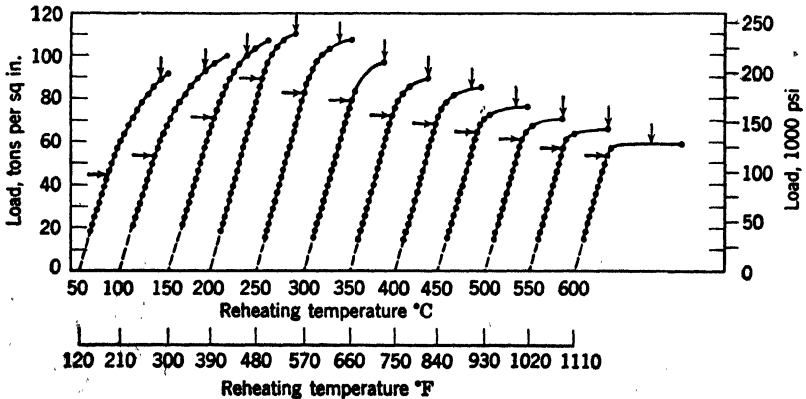


FIG. 122. Stress-strain curves for a cold-drawn 0.74% C steel, reheated 1 hr at the indicated temperatures. (Greaves)

work by Kaiser and Taylor. The change produced by cold work in type of stress-strain diagram is shown by Fig. 122 from Greaves.<sup>5</sup>

That time, as well as temperature, is involved is shown by Mann,<sup>6</sup> Fig. 123. There is no single fixed temperature for relief of residual stress.

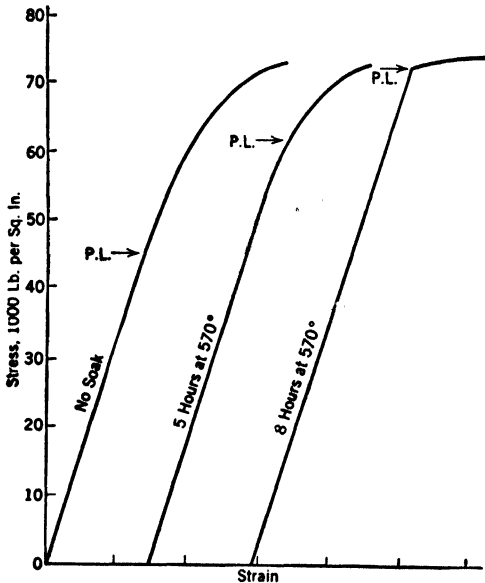


FIG. 123. Effect of tempering on the stress-strain diagram of cold-worked steel. (Mann<sup>6</sup>)

**Removal of Residual Stress by Tempering.** The tempering temperature after cold working for various steels will run from 400° on soft steels to 750° or a little above on the more highly alloyed steels. Temperatures of 475° for 0.22% C steel and 750° for a 0.26 C, 2.50 Ni, 0.55 Cr, 0.52% Mo steel were found suitable by Macrae.<sup>7</sup>

Tempering cold-drawn screw stock at 750° is considered to improve its machinability. Relief of residual stress after cold work has been experimentally studied by Stewart<sup>8</sup> for steels of 0.27 C, 0.43 Mn, 0.01% Si and 0.21 C, 1.44 Mn, 0.01% Si. The latter showed higher residual stress. Process annealing above 800° is required to produce plasticity enough to start much relief of stress even in 4 hr, and a temperature of 1200° for 1 hr is considered necessary to reduce residual stress to a negligible figure. The effect of welding procedure on residual stress is also discussed.

The severe deformation of cold work puts the material into an unstable state; it wants to go back to the stable state, and the more it has been disturbed, the less aid in the way of elevated temperature it needs to be able to go to a more stable condition. However, in low-C steels there is a peculiar phenomenon, by which, at a certain inter-

mediate but low degree of reduction, the tendency to grow large grains on recrystallization is very great; whereas, at smaller or larger degrees of reduction the recrystallization is to fine crystals which, at increasing temperatures grow, but grow slowly. Hence if, as in stamping an automobile fender, different parts of the piece are cold-worked to different degrees, and the piece then given a process anneal, the "critically strained" portions grow huge crystals, and the surface appearance suffers. The coarse-grained material is much more brittle than the fine-grained.

**Stretcher Strains.** A further complication arises in very low-C annealed steel during cold deformation in that, just after the yield strength is exceeded, the metal does not stretch uniformly, but some locations give before others, these small localities thin out, analogous to necking in a tensile test, and the surface is roughened. These surface markings are called "stretcher strains." Both the critical grain growth on annealing and the stretcher strains are obviated to a large degree when the C is raised to around 0.20%, but such steel is of course stronger and more difficult to form than softer lower-C steel, partly because it work-hardens more rapidly.

The cold-worked elongated grains naturally have directional properties; the transverse ductility and the impact resistance are lower than in the longitudinal direction (Fig. 74). More severe bending can be done with the fold longitudinal than with it transverse. Recrystallization minimizes this difference, but does not alter the distribution of the inclusions, and so directional differences still remain.

**Earing.** In speaking of directional properties, some comment should be made on "earing" and "fluting" of cold-worked stock. Cold rolling coaxes the crystal fragments to line up so that some particular direction within each crystal lies in the direction of rolling, the fragments tend to become "oriented," and, when recrystallization takes place in the process anneal, the crystals retain this orientation. Unless the yield strength and ability for deformation of each crystal are alike in all directions, and they usually are not, the array of oriented crystals is not indifferent to the direction of subsequent deformation. In drawing a cup from steel that has such orientation, four positions 90° apart will stretch farther than the rest, forming four "ears." To avoid this the degree of cold deformation and the process-annealing temperature have to be juggled, or a full anneal, into the austenite field, applied. "Fluting" is the term applied where a sheet, bent in a free bend, without restraint, bends

sharply at some points rather than in a uniform arc. It is met in forming up can body blanks.

**Fluting.** Fluting is present when the stress-strain diagram shows a sharp drop just beyond the yield point; it is not present when the

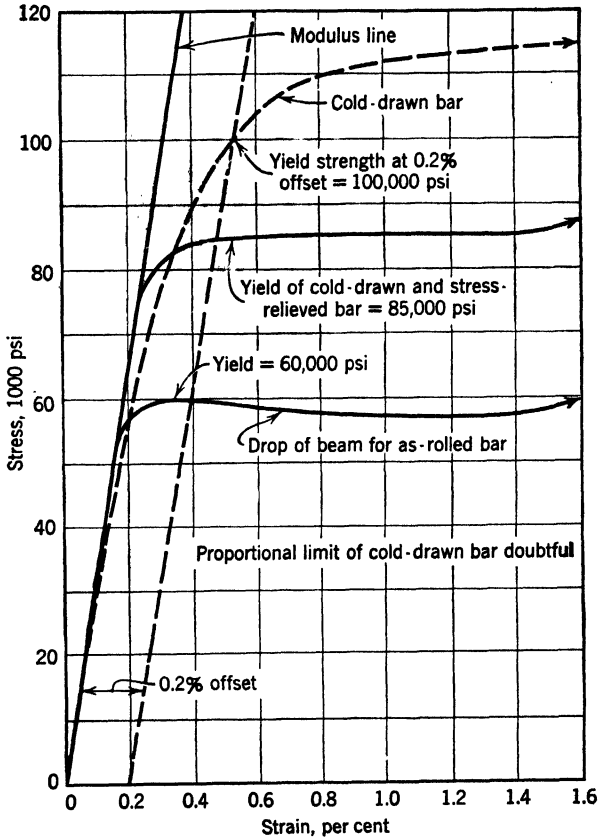


FIG. 124. Stress-strain curves for 1335 steel, as-rolled, cold-drawn  $\frac{1}{10}$ -in. draft, and cold-drawn and strain-annealed at  $1000^{\circ}\text{F}$ .

stress-strain diagram is uniformly curved, like the upper curve in Fig. 124 from Frye. Hence, in originally cold-worked sheet, fluting appears when the process anneal has been high enough to produce the other type of stress-strain diagram. Reintroduction of surface stress again gives the curved diagram. As Sisson and Clark<sup>10</sup> show, the degree of deformation and the process-annealing temperature can be juggled to avoid fluting, or a mere water quenching from the regular annealing temperature, which introduces internal stress,

may sometimes be used to advantage. The usual remedy, however, is to give a very light "skin pass," with reduction of only about 1 to 2% in thickness, which is sufficient to reintroduce enough stress at the surface to restore the stress-strain diagram to the curved type.

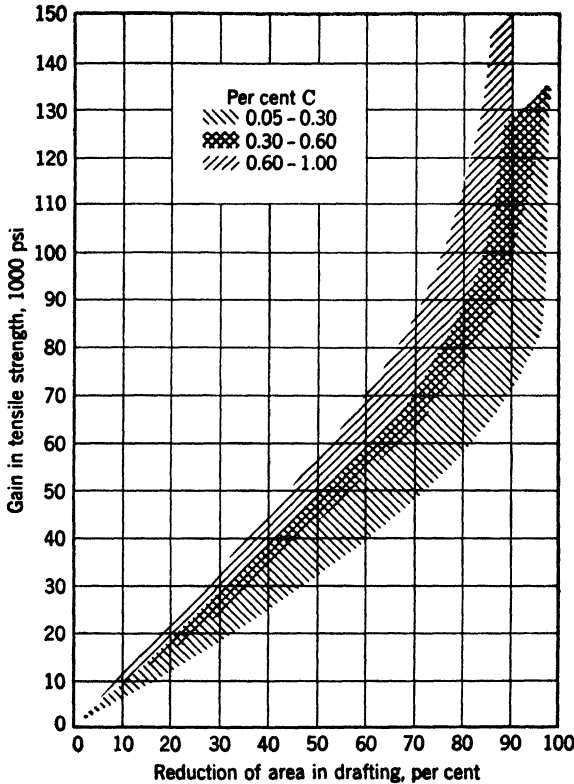


FIG. 125. Effect of cold drawing on the tensile strength of wire containing 0.05 to 1.00% C. (Legge)

Gold and Foulke<sup>43</sup> suggest a recrystallizing process anneal upon the surfaces only of black plate in a rolled temper such that it would otherwise flute. By utilizing the ability of very high-frequency induction heating to heat the surface only, and so quickly that insufficient heat penetrates by conduction beneath the surface skin to produce annealing, they say that a hard temper core with recrystallized surfaces can be produced which will not flute.

Skin passes (temper rolling) are also used to remedy propensity toward stretcher strains which likewise result from the stress-strain

curves having a sharp jog just above the yield point. Stretching<sup>11</sup> just past the yield point jog has a similar effect.

**Yield Strength.** Figures 122 and 123 show the increase in yield strength resulting from stress relief. A rise in yield strength due, in part, to stress relief, in low-C temper-rolled stock, occurs on storage at room temperature (see p. 244).

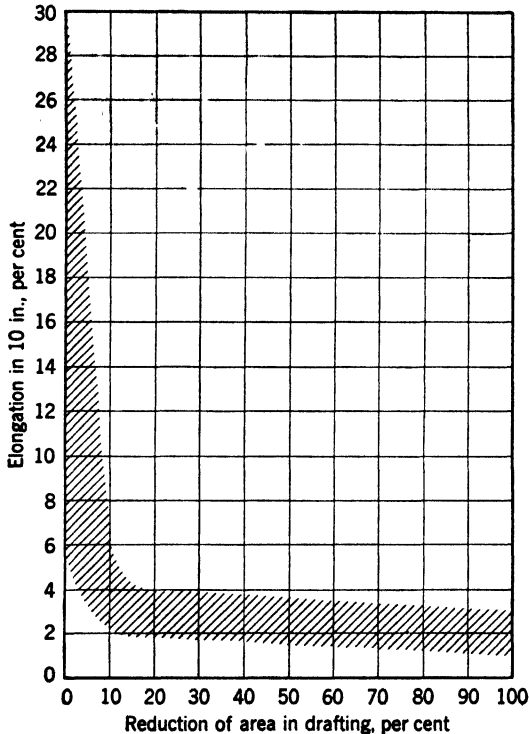


FIG. 126. Effect of cold drawing on the elongation of wire containing 0.05 to 1.00% C. (Legge)

The strength producible by cold work depends on the initial strength, that is, on the C content, and initial structure, but the *gain* in tensile strength is not greatly different. This is discussed by Hollomon.<sup>49</sup>

The elongation drops so rapidly with increase in cold work that the spread in ductility between low-C and high-C contents, above some 10% cold work, is small. These facts are shown in Figs. 125 and 126 from Legge.<sup>12</sup> Cold work raises the elastic strength consistently with this rapid drop in ductility.



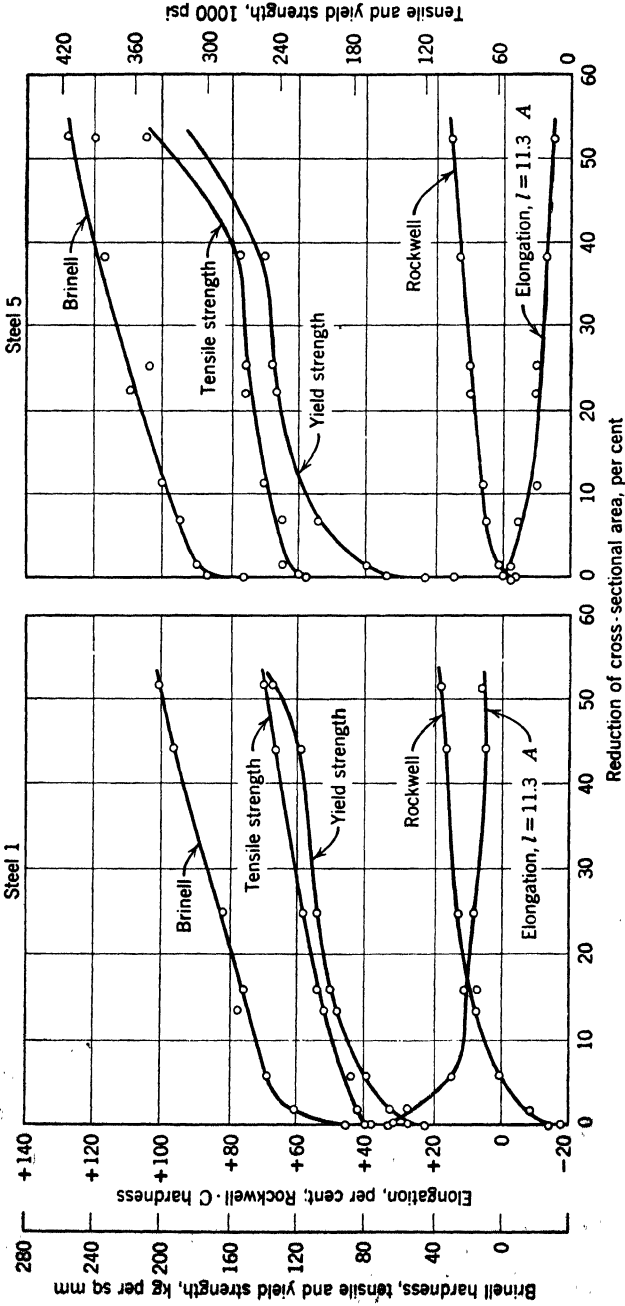


Fig. 127. Effect of the amount of reduction in cold rolling on the tensile properties of rod containing 0.12 to 0.61% C. (Greulich)

**Restoration of Ductility.** A process anneal can be used to restore some ductility without too much sacrifice in yield strength. This is illustrated by Balicki's<sup>13</sup> data for so-called "Armco iron," containing 0.01 C, 0.26% Mn, which, fully annealed, showed 42,000 tensile, 20,000 "plastic action," and 26% elongation.

Cold drawing to 84% reduction raises the strength of the initially normalized material, but gives low elongation.

Annealed at	84% REDUCTION					
	No Anneal	500°	785°	1020°	1125°	1345°
Ultimate strength, psi	89,000	85,000	80,000	60,000	50,000	47,000
Plastic action, psi	89,000	84,000	78,500	59,000	42,000	35,000
Elong., %	0	3	8	16	37	42

**The Stress-Strain Diagram.** The stress-strain diagram is slightly curved for cold-worked materials, owing to the internal strain, so

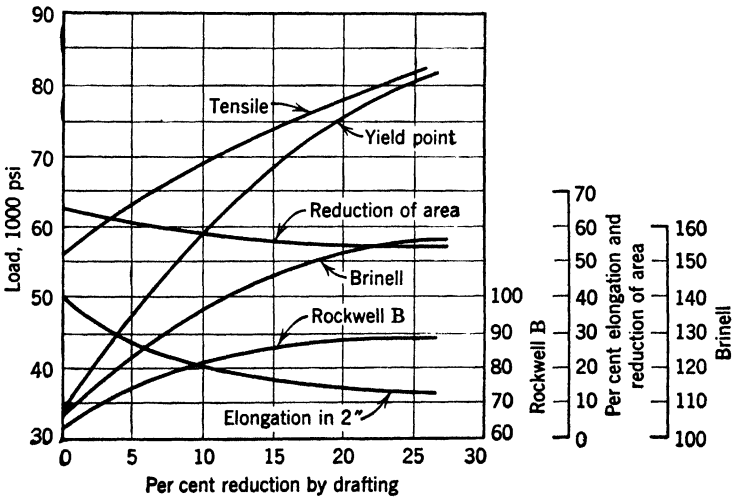


FIG. 128. Effect of drafting without reheating.

that the stress at which a certain degree of definite plastic action occurs is used by Balicki instead of yield strength.

The 785° anneal maintains high strength and imparts measurable ductility, whereas to soften and restore ductility ready for another severe cold deformation an anneal at 1125° or in that vicinity is needed. On the other hand, the anneal at around 500° imparts some relief from internal stress. The same sort of thing is shown for steel at 0.74 C in Fig. 122. The yield strength is raised at annealing temperatures up to around 400°.

Cold-rolled rods of 0.12 C, 0.40 Mn, 0% Si (steel No. 1) and of 0.61 C, 0.83 Mn, 0.34% Si (steel No. 5) show the data plotted in Fig. 127 from Greulich.<sup>14</sup> Intermediate-C content gives intermediate values. The behavior of C 1137 is shown in Fig. 128.

**Properties of Cold-Drawn Bars.** The general effects of variation in draft (amount of deformation) and of process-annealing temperature are shown in Fig. 129 from Landis.<sup>15</sup>

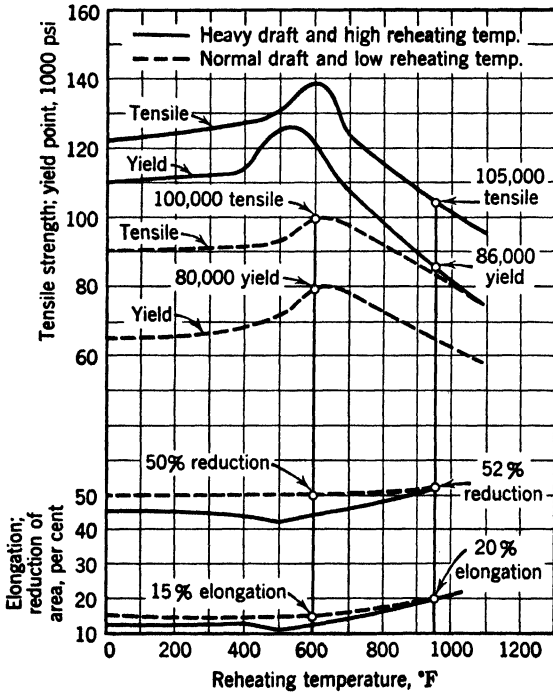


FIG. 129. Effect of drafting and reheating on properties of a typical cold-drawn C steel.

Cold drawing of bars is applied to steels from 0.10 to 0.45% C, and with Mn up to around 1.90%, with or without high S. Eutectoid, or even higher-C steels are cold-drawn into wire. Landis lists the average properties obtained in conventional cold drawing, over this range. He points out, however, that different drawing techniques affect properties, as shown in Table 8.

Moreover, there is a scatter in material processed from the same ingot, as shown in Table 9. Superimposed on this is a difference in original properties of hot-rolled bars, depending on mill practice, and

TABLE 8

PHYSICAL PROPERTY CHANGES RESULTING FROM VARIATIONS IN COLD-DRAWING PRACTICES ON A 0.40 C STEEL

	<i>Hot Rolled</i>	<i>Cold-Drawn with Varied Die Practices</i>		
Tensile strength	94,500	112,000	121,750	135,000
Yield point	63,500	97,000 *	113,500 *	129,000 *
Elongation in 2 in.	30.0	14.1	11.9	10.9
Reduction of area	54.1	50.6	45.5	44.6
Impact	60	38	34	32
Brinell hardness	207	223	235	255

\* Divider yield point.

TABLE 9

VARIATION IN PHYSICAL PROPERTIES OF HOT-ROLLED BARS FROM ONE INGOT OF C1141

<i>Physicals</i>	<i>Top</i>	<i>Middle</i>	<i>Bottom</i>
Tensile strength, psi	108,750	102,000	99,350
Yield point, psi	64,850	62,500	60,500
Elongation in 2 in.	22.5	26.9	28.2
Reduction of area	47.8	57.5	59.6
Izod, ft-lb	40	55	60
Grain size	2 to 5	4	4 to 5
Deep etch	OK—slight pattern and pits	Good	Excellent

TABLE 10

MINIMUM PHYSICAL PROPERTIES FOR BARS COLD-DRAWN WITH HEAVIER THAN NORMAL DRAFTS AND WITH SUBSEQUENT FURNACE TREATMENT

<i>Analysis</i>	<i>Size of Rounds</i>	<i>Tensile Strength, Psi</i>	<i>Yield Point, Psi</i>	<i>Elong. in 2 In.</i>	<i>Reduction Area</i>	<i>Brinell Hardness</i>
C1045	½ in.	135,000	105,000	8.0%	30.0%	240 min.
C1144	3	120,000	90,000	10.0	30.0	
C1050	¾	120,000	100,000	10.0	30.0	255 to 302
C1050	1½	120,000	115,000	10.0	30.0	241 to 289
C1040	1	115,000	95,000	12.0	40.0	
C1137	⅞		90,000			
C1045	2		95,000			
C1141	1¼		100,000			
C1137	2½		80,000			
C1045	1½	120,000	100,000	12.0	40.0	
C1141	2¼	120,000	95,000	10.0	30.0	
C1144	2		100,000			
C1045	1½	115,000	90,000	10.0	35.0	

bar size. Since a given amount of cold work steps up the strength from its original level, differences in the original level remain after cold drawing.

Optimum mechanical properties are produced by rather heavy drafts followed by the proper process anneal. For the 0.35–0.45% group Landis gives the data in Table 10. Of course, it must be remembered that the ductility figures cited by Landis are for longitudinal specimens.

**“Warm Work.”** While the explanation may not be quite so simple, plastic deformation applied in the range of temperature used in process annealing, to straighten out the stress–strain diagram and increase the yield strength, appears to accomplish the stress relief in the same operation that produces the deformation. Dunn<sup>60</sup> describes how a rolling reduction of  $5\frac{1}{3}\%$  applied to oil-well casing of 0.41 C, 0.18 S, 1.06% Mn steel raised the yield strength and lowered the elongation variously, depending on the temperature of deformation, as shown in Table 11.

TABLE 11

	Yield Strength, Psi	Elong. % in 2 In.
Hot-rolled, no working	58,500	26½
5⅓% reduction by working applied at:		
Room temperature	92,000	24
500°	112,500	16
650°	108,500	16½
850°	98,000	19

**Depth of Strengthening.** Cold drawing at usual reductions strengthens even large bars to the center; hence cold drawing affords a means of securing like properties at center and surface in sizes that would not harden to the center on quenching. With only 15% reduction a steel initially 83 Rb is increased to around 91 clear to the center, with the material near the surface only a trace harder. Frye<sup>9</sup> cites the case of 1020 steel in  $2\frac{1}{2}$ -in. rod, cold-drawn by a  $\frac{1}{16}$ -in. draft, which gave in 0.505-in. bars taken from edge to center:

Annealed	Yield	Tensile	Cold-Drawn	Yield	Tensile
Edge	41,500	70,000	Edge	65,000	75,000
Center	35,000	66,500	Center	62,500	72,000

No heat treatment would give the center of a 2½-in. bar of 1020 the yield strength produced by cold drawing.

The cold-drawn material has a curved stress-strain diagram, the "proportional limit" being very low and there being no marked change in slope at the deformation taken, by definition, as the yield strength. Stress relieving at 700–750° straightens out the diagram greatly, raising the "proportional limit" and increasing both the

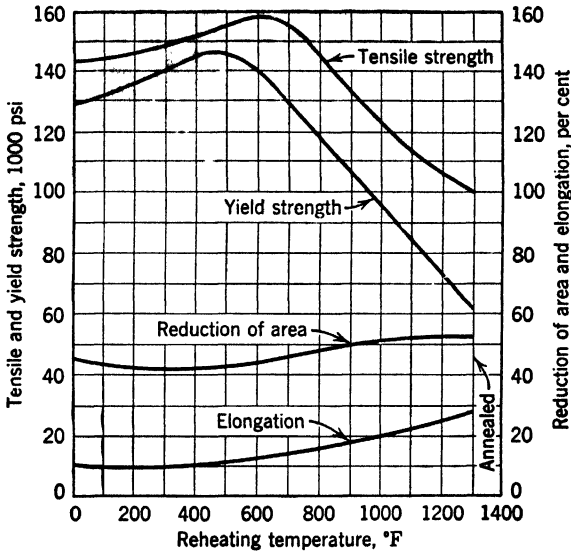


FIG. 130. Average effect of strain annealing cold-drawn bars at various temperatures. (Courtesy LaSalle Steel Co.)

yield and tensile strengths with only slight change in ductility. As the annealing temperature is raised further, strength decreases and ductility rises, as Frye shows in Fig. 130.

**Variability in Work Hardenability.** As Fig. 121 shows, the greater the degree of cold reduction, the lower the temperature needed to produce recrystallization. Different heats of steel respond differently to cold work, some work-harden more readily than others, as Graham and Work<sup>16</sup> show, in Figs. 131 and 132. This is doubtless related to the degree of deoxidation. Graham and Work evaluate work hardenability by drawing a tapered bar through a die, so as to produce different degrees of reduction along its length. Notches are cut at distances corresponding to known degrees of reduction, and the bar is fractured by impact at each notch. They point out that

the test is not applicable to rimmed steel whose core is different from its skin. Only small degrees of reduction are shown in Figs. 131 and 132. It may be seen that different types of deformation have different effects, even though the degree of deformation is the same.

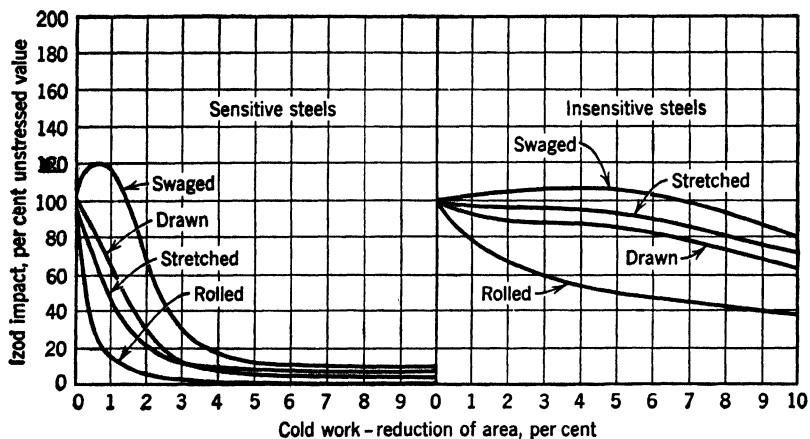


Fig. 131. Effects of various types of cold work on sensitivity.

This is corroborated by Kaiser and Taylor,<sup>17</sup> who show also that, under the same permanent deformation, grain growth on process annealing varies with the type of deformation.

**Deoxidation Practice.** Figure 132 shows that different heats work-harden differently. The effects of differences in deoxidation are

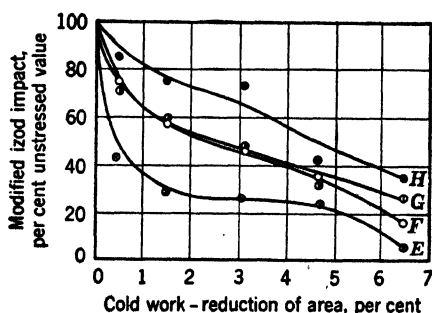


Fig. 132. Sensitivity characteristics of 0.60 C, 1.00% Mn steels.

brought out by Work and Enzian.<sup>45</sup> In stock for deep drawing, grain size is important, and the process-annealing temperature has to be adjusted so that grain growth will proceed to the desired size, but not beyond it. The variability in the relation between

degree of deformation and recrystallization temperature in process annealing was shown in Fig. 121. Wallace and Rickett<sup>18</sup> show the greatly different response to process annealing of three steels of the compositions indicated in Table 12.

TABLE 12

	C	Mn	P	S	Si	Ferrite Grain, Size as Rolled
A Al-killed	0.15	0.44	0.02	0.05	0.04	7
B Si-killed and re-phosphorized	0.11	0.31	0.05	0.05	0.10	7
C Rimmed (core)	0.03	0.09	0.01	0.06	tr	6
Rimmed (rim)	0.02	0.08	tr	0.02	tr	6

These were cold-drawn from 1-in. hot-rolled rod to 0.433 in. diameter, that is, 81% reduction.

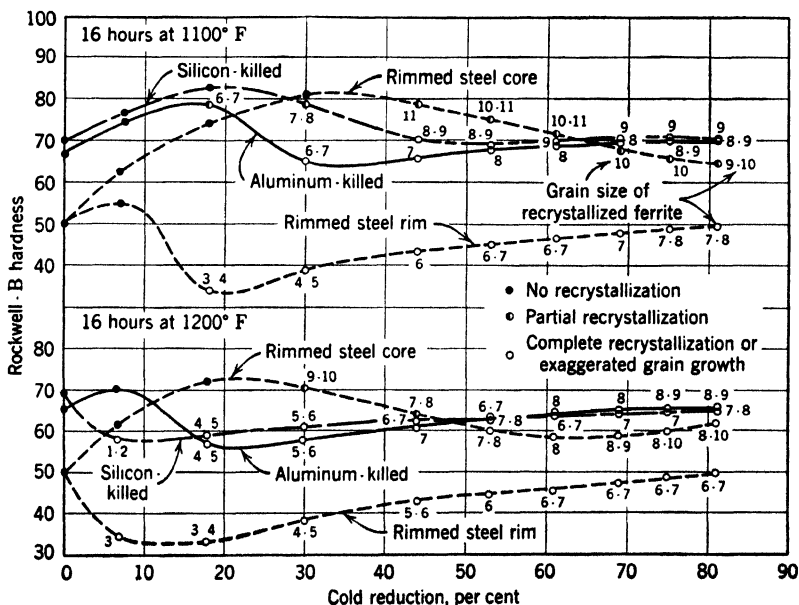


FIG. 133. Effect of type of steel and amount of cold reduction on response to annealing at 1100 to 1200°F for 16 hr.

Figure 133 shows how individually the steels behaved.

Thus, for each heat of steel, as well as for steels of different composition, the degree of reduction and the process-annealing time and



temperature will be varied according to the behavior of the particular lot, though there is a general pattern of behavior followed by all steels.

**The Time-Temperature Relation.** Balicki<sup>13</sup> points out, as do Wallace and Rickett,<sup>18</sup> that the time required for material softening in a process anneal at high temperatures is short, a matter of minutes. The time-temperature relations in process annealing have

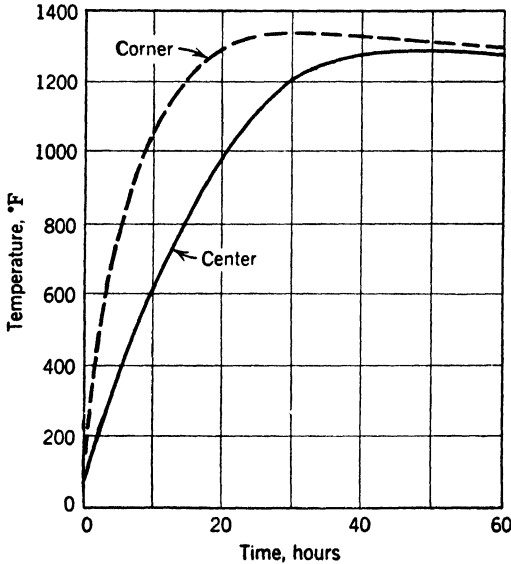


FIG. 134. Temperature near corner and at center of pile of sheets during annealing.

not been thoroughly worked out. In box annealing practice on large charges, often 25 tons or more, where heat flow to the center of a coil or a pack of sheets is very slow, the temperature is held at a level that will produce just the required softening in a wide range of times, so that the outside, heated longer, will not be much softer than the inside, which is at the maximum temperature for a much shorter period; hence ordinary process annealing has usually been discussed merely in terms of temperature, rather than of temperature and time. Wallace and Rickett<sup>18</sup> show that at higher temperatures, 15 min will accomplish the softening that requires many hours at the usual temperatures. The necessity for long time at the usual temperatures to get heat penetration to the center of a pack (Fig. 134) is brought out by Johnston.<sup>19</sup>

That only short times at high temperatures are necessary for considerable softening is brought out by experience with process annealing of steel cartridge cases to be discussed later.

**Cold-Drawn Free-Machining Steels.** Bar stock of steels high in S is frequently cold-drawn for increase in machinability and to increase strength. In such steels, directional differences are huge.

**Compressive Yield Strength.** Marked directional differences are met in cold-worked steel. Most of the quantitative data comparing tensile and compressive stress-strain diagrams are for austenitic stainless steel, but there are qualitative indications that much the same situation exists with cold-worked ferritic steels. When no stress-relief anneal and no period of aging are used, the "Bauschinger effect" is noted; that is, although the yield and tensile strengths are increased when the tensile test is made with the stress applied in the same direction in which the cold working was done, in a compression test, the compressive yield stress is nearly zero, and the compressive stress-strain curve is curved from the origin; in other words, the material has apparently been strengthened in but one direction. A stress-relief anneal, insufficient to soften the material, gives a straight portion to the compressive stress-strain, as well as increases the straight portion of the tensile stress-strain curve, but the compressive yield strength is considerably lower than the tensile yield strength. However, on specimens stressed transversely to the direction of working, the reverse is found, the compressive yield strength is higher than the tensile yield strength.

In some cold-worked objects, notably cold-drawn wire, the service stresses come in the direction of the cold working, and the behavior in compression is immaterial. In many other objects the factor of safety in design is such that the directional differences are not of moment. In the service of certain aircraft parts, buckling is the primary concern, and so, if the compressive yield is low, plastic deformation will start on the compression side of a member when the compressive yield strength is exceeded, although the stress on the tension side is still within the elastic range. Hence equality of tensile and compressive yield strengths, such as is found in heat-treated steels, is desired, as well as that both yield strengths be high.

It is obvious that the absolute and the relative positions of the yield strengths can be influenced by variation in the reduction applied in cold working, by working in more than one direction, as by cross-rolling, and by the stress-relief anneal. Published information on these matters appears to be wholly lacking for ferritic steels.

Properties cited<sup>20</sup> for 1-in. round bars cold drawn with  $\frac{1}{16}$ -in. draft are given in Table 13.

TABLE 13

Grade	Tensile Strength, Psi	Yield Strength, Psi	Elong. in 2 In., %	Reduction of Area, %	Brinell Hardness
C-1113	76,000	67,000	20	50	159
C-1115	78,000	69,000	20	52	163
C-1116	80,000	70,000	19	48	167
C-1117	83,000	74,000	21	52	170
C-1118	86,000	76,000	20	50	174
C-1217	88,000	78,000	15	45	187

Considerable variation will occur from heat to heat and with size of bar, since the amount of drafting varies with the size. It is stated<sup>20</sup> that these open-hearth relatively high-Mn very high-S steels are "inherently more resistant to shock than, and not brittle like, Bessemer." This statement may be correct by comparison with high-N Bessemer steel, but is hardly to be taken at face value.

**Transverse Properties.** The properties of cold-drawn steels usually listed are those determined in the longitudinal direction only. Few data are available on the transverse properties of cold-worked steel. In order to supply such information, tests were made at Battelle for inclusion here.

A free-machining steel of 0.45 C, 1.50 Mn, 0.20 S, 0.02% P was cold-drawn from  $2\frac{1}{16}$  in. to  $2\frac{5}{8}$  in., stress-relieved, and straightened. Tensile and impact tests were made on longitudinal and transverse specimens. The longitudinal tensile tests on regular 0.505 in. diameter 2 in. gage length and on subsize bars, 0.188 in. diameter, 0.75 in. gage length, agreed well. Their average is compared below with the average of closely agreeing tests on subsize transverse specimens:

	Tensile, Psi	Yield, Psi	Elong., %	R.A., %	Standard-Size Charpy Impact, ft-lb	
					Keyhole Notch	V Notch
Longitudinal	125,000-129,000	95,000-110,000	13-13½	32½-34½	10½-12½	6½-8½
Transverse	106,000-107,000	91,500-96,000	4	2-3	1½	2

Without a great change in the yield strength, the ductility and impact resistance in the transverse direction fall off vastly. Some of the energy absorbed in the impact test is used in knocking the

bar off from the support, and so something has to be shaved off from even these low values.

In a further series of tests, made at Battelle for inclusion here, X1345 (0.45 C, 1.50 Mn, 0.25% S) was less severely drawn (2½ in. to 2¼ in.), and subsize tensile and standard V-notch Charpy tests made both on the hot-rolled and cold-drawn bars (Table 14).

TABLE 14

	<i>Tensile,</i> <i>Psi</i>	<i>Yield</i> <i>(2% Set),</i> <i>Psi</i>	<i>Elong.</i> <i>in ¾</i> <i>in., %</i>	<i>R.A.,</i> <i>%</i>	<i>Charpy V,</i> <i>ft-lb</i>
Hot-rolled longitudinal	102,500	62,000	26½	46½	41 to 70
Hot-rolled transverse	85,000	57,000	6	5	2 to 2½
Cold-drawn longitudinal	105,000	73,000	20	45	62 to 92 *
Cold-drawn transverse	86,000	74,000	4	4½	1½ to 2

\* These values are not comparable with the others, since the specimens did not fracture in the usual fashion. Instead of a fracture progressing in the direction of the apex of the notch, the specimens started to split lengthwise, and, when the split had progressed far enough to decrease the stiffness of the specimen sufficiently, it bent and was forced off the supports without complete fracture. This woody splitting in itself indicates extreme directional weakness.

Similar tests were made on B1112 (0.10 C, 0.75 Mn, 0.20% S) drawn from 2½ in. to 2¼ in. (Table 15).

TABLE 15

	<i>Tensile,</i> <i>Psi</i>	<i>Yield</i> <i>(2% Set),</i> <i>Psi</i>	<i>Elong.</i> <i>in ¾</i> <i>in., %</i>	<i>R.A.,</i> <i>%</i>	<i>Charpy V,</i> <i>ft-lb</i>
Hot-rolled longitudinal	67,000	40,000	37	60	16 to 19½
Hot-rolled transverse	62,500	36,000	22½	24½	8 to 9½
Cold-drawn longitudinal	78,500	68,000	20½	53	3 to 3½
Cold-drawn transverse	75,000	60,000	9	29	3½ to 5

Further, such tests were carried out on C1019 (0.19 C, 0.85 Mn, 0.05% S maximum) drawn from 2¼ in. to 2½ in. (Table 16).

TABLE 16

	<i>Tensile,</i> <i>Psi</i>	<i>Yield</i> <i>(2% Set),</i> <i>Psi</i>	<i>Elong.</i> <i>in ¾</i> <i>in., %</i>	<i>R.A.,</i> <i>%</i>	<i>Charpy V,</i> <i>ft-lb</i>
Hot-rolled longitudinal	64,000	36,000	43	66	75 to 96
Hot-rolled transverse	62,000	32,500	36	54	33½ to 37½
Cold-drawn longitudinal	71,500	61,000	26	64½	17 to 38
Cold-drawn transverse	68,000	50,000	16	50	7 to 12

The toughness of the high-S steels in the transverse direction is low even in the hot-rolled condition. Despite the rather high longitudinal reduction of area, the toughness in the transverse direction is lower in the cold-drawn high-S steels than in malleable iron. Such cold-drawn free-machining steel is almost as brittle as cast iron to stresses transverse to the direction of rolling. Yet the ductility figures normally reported only for longitudinal specimens would indicate toughness, and the longitudinal impact values, although low, would nevertheless lead to the belief that some toughness does exist.

Much such steel is used successfully for parts that are subjected to low service stresses, when stress-raising notches are avoided, or the service does not involve impact; that is, when the real criterion of serviceability is the static yield strength, such cold-drawn steel is good steel, but it is wise to watch where you put it. Williams<sup>21</sup> says, "Such steel may not be good for spline shafts where concentrated stresses may coincide with inclusions," and Howe<sup>22</sup> says, "One is reluctant to believe that this material has been used for the steering posts of motor cars. Such practice should be a felony."

Riegel<sup>23</sup> remarks, "The deceased occupants of the wrecked vehicle on the bottom of the mountain chasm will never know that the gyp steering knuckle made of brittle screw stock was nondependable." The fatigue resistance of such steel, cold-drawn to relatively high hardness, would presumably be poor.

These comments should not apply to uses where the accident insurance of some toughness is not needed, or where fatigue is not a factor, and there are many such. But our usual ignorance about, and ignoring of, transverse properties in such bars, and the properties in the through-plate direction in plates (see p. 118), are no credit to the metallurgist or the design engineer.

As usual, we cannot eat our cake and have it too. Piling in lots of S helps machinability, but depreciates transverse toughness. It is necessary to adopt a compromise when transverse toughness counts. Holding S at around 0.05% often gives acceptable toughness and sufficiently improved machinability to be worth while.

**Cartridge Cases.** One of the most interesting and important applications of cold drawing is in steel cartridge cases for fixed ammunition. Sizes of interest range from 22 cal, which were sometimes made of steel commercially before the war, through 30-cal, 37-mm, 105-mm, and 3-in. antiaircraft shells. Each size is a distinct manu-

facturing and processing problem by itself, to be worked out individually.

The finished case has a heavy base with a central projection to hold the primer, a long body with some taper, and a smaller mouth into which the projectile is crimped. The stock starts out as a disk, then is cupped, and then the cup lengthened into a heavy-bottomed cylinder in several successive draws which decrease the wall thickness some 33% to 45% at each step. The steel is annealed between each draw to restore formability. The base is then formed into shape by a cold-forging operation. Next the cylinder is tapered, then the mouth is locally annealed to permit its final shaping, and finally a stress-relief tempering is given.

The base must be strong to resist the powder pressure. The walls must have enough elastic range to permit expansion of the case during firing, to take up the clearance between case and chamber, and then to allow the case to contract when the pressure is released, so that the empty case can be extracted without sticking. The low modulus of elasticity of the usual brass accomplishes this, even though the yield strength is not high.

The steel case must have a high yield strength so that its expansion and contraction are wholly elastic. German practice<sup>28</sup> used steel of 0.10 to 0.20% C sometimes pearlitic at the start, sometimes partly or wholly spheroidized, with process anneals between draws, and was content with a yield strength of 75,000–90,000 psi, which was on the ragged edge for satisfactory extraction.

American practice used a C range of 0.20–0.30%, and the steel was Al killed. The cases have 115,000–125,000 psi yield strength. If the C is chosen on the low end of this range the blank can be pearlitic at the start, at 0.25% or more C it is usually fully spheroidized. Sometimes a full anneal at 1650° is given between draws;<sup>24–27</sup> sometimes a 1250° process anneal. To retain strength at the base, after the base has been cold-formed, the anneal prior to tapering is local, either by immersing to the proper depth in a salt bath, or by local induction heating, to give a short process anneal, ranging from 2 min at 1100° to 3 sec at 1250°. The final mouth anneal is similarly localized and is usually a 1250° process anneal. A final 650° or 725° stress-relief anneal is given the whole case. In the 3-in. size the base and side walls need greater strength and are locally heated and drastically quenched with water spray which on

this thin section is effective at least to produce lower bainite, then being tempered to the right degree by the 725° final tempering.

The completed case then has either a cold-worked or a quenched and tempered base, with the side wall and neck, more severely cold worked, softened back just enough by the final process anneal.

There is a particular gradation of hardness from base to mouth needed for each size and design of case.

Prior to the experience with steel cartridge cases it had been supposed that such severe draws on steel of 0.30% C and such final strengths were unobtainable save at the cost of prohibitive breakage in drawing. The use of controlled annealing, coupled with improved drawing practice as to dies and lubricants, brought the breakage with 0.30% steel down to about what used to be met with 0.10% C steel under similar draws and gave the necessary higher strength.

It is particularly interesting that the ring near the base is softened where the local heating for quenching just fails to produce austenite, but removes the hardness provided by cold work, and another just back of the mouth, softened by the mouth anneal but not rehardened by crimping, do not lead to local failure. These tiny weak rings are supported by the adjacent stronger material.

**Aging.** Most discussions of cold-worked steel include comment on "aging." Stress-relief and process annealing both depend on the urge of the mashed-up crystals to return to a more orderly arrangement, through infinitesimal plastic flow or creep. The time factor as well as temperature is involved. Recalling the low proportional limit of the curved stress-strain diagram, it is not unexpected that some readjustment should occur on long standing even at room temperature. Frye<sup>9</sup> shows that, in cold-drawn bars from low-C 1112 steel, yield and tensile strength increase on aging, and the ductility simultaneously improves, just as occurs more quickly on stress-relief annealing. The higher-C steel 1135 changes much less. As is pointed out by Sisco,<sup>29</sup> aging behavior is quite erratic; just as response to cold work varies from heat to heat, so does aging behavior.

Cold-rolled low-C steels may become somewhat embrittled on aging, instead of improving in ductility as did Frye's 1112, but this is probably due to the entrance of another factor, "precipitation hardening." To differentiate between the two factors, that related to cold work alone is called "strain aging," that due to precipitation hardening, "quench aging." The latter is discussed on p. 256. Both

types of aging are less pronounced and generally negligible in steels of the C content usually employed for quenching and tempering.

**Patenting.** Just as the C content sets the initial strength level upon which added strength is built by cold working, so the initial structure in a steel of given C content sets its initial strength level and affects the response to cold work. A preliminary heat treatment from above the critical is therefore often applied prior to cold working.

This is especially applicable in the case of cold-drawn high-C wire for piano wire, and, in larger sizes, for suspension-bridge cables. The austenite is first coarsened and thus made more sluggish by heating to a high temperature, and then either air-cooled, or, preferably, "quenched" into a bath of molten lead held at a desired temperature not very far below the critical, in order to facilitate transformation to fine pearlite. This is an early use of the process now called "austempering" (pp. 387, 417). The fine pearlite structure is stronger than a coarser one, and the final product has greater strength than if the cold work were applied to a coarser structure. Godfrey<sup>30</sup> studied three steels of 0.75–0.80 C, 0.50–0.65% Mn, lead-patented at different temperatures or air-patented, then similarly cold-drawn. Lead patenting at 900° gave, after cold reduction, wire of 190,000 tensile, 7–7½% elongation, 60–65% R.A. Results from patenting at 1000° were a bit lower in tensile and R.A. and more erratic. Among the three, the 1100° patenting was notably inferior, and the air-patented wire only had 165,000 tensile with 7% elongation, and 53% R.A. On the basis of true strain at fracture, the 190,000 tensile strength of the cold-drawn wire is equivalent to 300,000 psi. The work hardening in cold drawing increased with the strength and fineness of the patented structure.

Oddly, the still finer structure obtained by martensitic quenching and tempering is not considered to behave so well in cold drawing as does fine pearlite. Hollomon<sup>31</sup> presents data comparing work-hardening behavior (determined in tensile testing by "true stress-strain curves") which seems to throw some doubt on this belief. Hollomon's work again shows the dependence of response to cold work on C content and initial strength.

By increasing the Mn, the steel can be made sluggish enough so that in air cooling the pearlite transformation temperature is lowered enough to produce fine pearlite, but, for plain C steels, the lead bath is considered more satisfactory. Epstein<sup>32</sup> and Sisco<sup>29</sup> give discussions of the patenting process. Thompson<sup>33</sup> gives a brief



description of cold work, process annealing, and patenting of cold-drawn wire, with micrographs of structures.

Before passing to precipitation hardening, another type of "annealing" should be discussed. This is included in this chapter, because, like process annealing, it involves no transformations in the steel.

**Heat Treatment to Remove Effects of Hydrogen.** "Flakes" in both C and alloy steels, and "shatter cracks" in rails (see Fig. 98) are ascribed to a variety of causes, but the consensus of opinion and the indications of many experiments are that hydrogen is the chief culprit.

When the steel is cast, according to the usually accepted theory, a very tiny amount of hydrogen is held in solution in the melt, as atoms, not then existing as a gas molecule. As the solid metal cools, the H atom comes out of solution and tries to diffuse out. If it succeeds, no harm is done. But, if the diffusing H finds its way into an internal void, such as one resulting from solidification shrinkage, that is, distributed pipe, two atoms combine to form a  $H_2$  molecule. As the molecules collect, the trapped gas exerts terrific pressure, higher than the tensile strength of the steel, and rips the steel apart in a small area around the void. These rips do not occur close to the surface, since the H atom has a better chance to escape from the surface without joining up with another, and so the cracks and splits are internal.

In hard steels these cracks do not always weld up on rolling or forging and remain as internal stress raisers. In rails, transverse fissures are believed to grow from the shatter cracks, under the influence of repeated stress. Since the fissure grows from the inside, it is not visible, and, before the installation of "detector cars" which record the electrical resistivity of the rail and thus make possible the removal of a fissured rail before fracture, many serious accidents occurred.

The escape of H from the cooling steel is easiest, not at an extremely high temperature, but at only moderately elevated temperatures. Hence, if the rail or other object is put, preferably while still hot from rolling or forging, into a low-temperature furnace, or into a well-insulated cooling pit, so that it cools very slowly from near the rolling temperature down to  $300^\circ$  or even to  $120^\circ$ , the escape is facilitated.<sup>34</sup> Opinions vary on to how low a temperature the slow cooling must be carried down.

Rails have regularly been slow-cooled in this way for a decade, and the fissure and accident records prove the virtue of the practice.

Kingsley<sup>35</sup> reports that in large billets of alloy steels slow cooling is not enough to avoid flaking, for the thermal stresses set up owing to temperature differences between outside and inside, plus the stress due to transformation at  $A_{r1}$ , plus the stress from H, all acting together, have to be considered. He reports that the steel should be rapidly cooled below  $A_{r1}$  and allowed to equalize in temperature so that all of the billet is below  $A_{r1}$ , but not allowed to cool far enough to allow the H to precipitate. From this temperature (1200–700°, depending on the type of steel), it is reheated to just below  $A_{c1}$ , furnace-cooled to 1200°, and then air-cooled. The H is retained, but the theory is not to let the different stresses be generated at the same time, so that, in normalized, stress-relieved billets, the stress from H will be insufficient to rupture the steel. Kingsley gives the temperature cycles for common alloy steels in various-sized billets, and describes a 345-ft-long tunnel kiln with three different temperature zones through which the billets are passed.

**No One Schedule Universally Applicable.** Andrew and co-workers<sup>44</sup> made an extensive investigation of escape of  $H_2$  at different temperatures, using technique similar to that employed in obtaining "S curves." They showed that the gas diffuses slowly from austenite, and the more slowly, the lower the temperature. At the transformation temperature, where gamma and alpha iron coexist, the gas concentrates in the gamma regions. It diffuses more rapidly from alpha than from gamma. Low-alloy contents do not alter the diffusion, but composition does alter the temperature of the gamma to alpha transformation.

No general pattern of behavior was found for all steels; the optimum temperature for escape of  $H_2$  varied all over the lot. The size of the piece of steel greatly affects the rate of escape: it is very slow from large sections. The relation of H to hairline cracking was studied, and cracking was found very difficult to avoid in many steels. The work indicates that no particular cooling schedule will serve for all steels, and that each steel needs to be studied as a particular case. Much of the evidence, however, is not incompatible with the principles suggested by Kingsley.

Gross and coworkers<sup>48</sup> found flakes or "fisheyes," ascribed to H, in as-rolled plate of 0.14 C, 1.30 Mn, 0.25 Si, 0.033% Ti, which disappeared after 8 hr at 400° or 32 days at room temperature.

Lorig and Samuels<sup>47</sup> report that the presence of H in malleable iron materially influences the behavior during heat treatment. Holding at 600° for 6 or 7 hr allows escape of H, and, on subsequent heat treatment for malleablization, the temper C is finer and better distributed than when H has not been removed. Vennerholm<sup>48</sup> comments on the commercial use of the dehydrogenating treatment.

There has been much discussion of the causes of flakes in ingots and heavy forgings of alloy steel, and of whether or not the flakes in ingots will weld up in rolling. Tovenets,<sup>36</sup> working with 7½-in.-diameter by 12-in.-long slugs of 0.33 C, 0.40 Si, 0.40 Mn, 0.78 Cr, 2.84 Ni, 0.30% Mo steel, concluded that flakes form below 300°, and the presence of martensite renders steel susceptible to flaking. He therefore suggests holding above the Ms point or reheating without having cooled below the Ms point, to give time for transformation to austenite to nonmartensitic products, before slow cooling to permit escape of hydrogen.

Slow cooling, or a prompt low-temperature reheating after cooling, is often applied to steel castings, or at least to test bars which are to be representative of the properties of the steel after hydrogen has escaped. The ductility of reheated test bars is often appreciably greater than that of a fresh unheated bar. Hydrogen seems to be able to escape in a month or so at room temperature, from small sections, and flakes or cracks ascribable to hydrogen are seldom found in light steel castings. However, in heavy sections, neither storage nor low reheating seems to allow effective escape of hydrogen and establishment of the ductility that would have been shown had H not been present.

**Limitations of Analysis.** The amount of H concerned is very small on a weight basis and very difficult to determine analytically,<sup>37</sup> but its volume when combined as molecular gas is often equal to or greater than the volume of the steel. Even were it simple to determine the hydrogen content with exactness, correlation of hydrogen content with impaired mechanical properties would require starting with the melt itself, since the escape of the gas from the solid metal at room temperature is erratic. The amount left in and found in the steel at the time the analysis is made thus bears no definite relation to its original content nor to the amount of harm resulting from its presence.

**Hydrogen in Welds.** The evil effect of H is perhaps most marked in welds. In gas welding the H of the hydrocarbon burns to water vapor and the water vapor reacts with the hot steel to form Fe

oxide plus  $H_2$ , some of the  $H_2$  dissolving in the steel. In electric welding with welding rods coated to give stability to the arc and provide fluxes for slagging off oxide, moisture held as water of crystallization in some constituent of the coating, or merely adsorbed in the coating, or the H in the cellulose of cellulose-coated rods, is similarly available to injure the weld. Hydrogen is increasingly being held responsible for cracks in and around welds, which cracks were previously ascribed wholly to the production of hard zones martensite or lower bainite resulting from transformation of austenite. Strict limits were set to hardness in the heat-affected zone. It is becoming evident that the mere presence of a thin hard zone, backed up on both sides by softer tougher structures, is not necessarily harmful, and that cracking which is harmful can more easily be avoided if H is not present.

Evidence is accumulating that some types of weld cracking are due to the simultaneous presence of martensite and H, the latter being responsible for the cracking of the former. Avoiding martensite by use of a nonhardenable steel or avoiding H by selection of the welding-rod coating helps to prevent cracking. The latter expedient opens the way to welding of strong steels previously considered unweldable.

Further evidence as to the presence and effect of H in welds is supplied by Zapffe and Sims,<sup>38</sup> Reeve,<sup>39</sup> Rollason,<sup>40</sup> Sloman and co-workers,<sup>41</sup> Herres,<sup>51</sup> and Voldrich.<sup>52</sup>

**Hydrogen from Controlled Atmospheres.** Liedholm<sup>53</sup> reports experiences in quench-cracking of alloy steel blades for aircraft propellers, heated for quenching in a protective atmosphere high in hydrogen (35%), which indicated that hydrogen was probably responsible for the cracking.

**Pickling Embrittlement.** Hydrogen is evolved as ions during the attack of certain acids on Fe in pickling. Two H ions readily combine to form relatively innocuous molecular  $H_2$ , but as long as the ion is uncombined it is in an active state. Since the ions are produced directly at the metal surface, some of them enter the metal. As soon as the metal is removed from the acid the H in the metal begins to diffuse out, as well as to try to diffuse further in. Since, in a brief pickling operation the H does not have time to penetrate far, only the surface layer is originally affected. On room-temperature storage, or mild heating, as in boiling water, the bulk of the H escapes from the surface layer without much deep penetration. Rinsing off with acid with boiling water therefore helps. While the

surface is loaded with H, the surface is brittle, and a crack starting in the brittle layer may readily progress through the unaffected interior. Forming operations cannot be performed on steel in this brittle state without breakage, and so either storage or a mild heating must intervene before such steel is formed.

Certain impurities in acids facilitate the absorption of H, while some "inhibitors" are adsorbed on the steel surface as soon as it has been cleaned by the acid, preventing further attack and production of H, and hence mitigating embrittlement. Compare Kehl and Offenbauer.<sup>42</sup> Although pickling embrittlement is a serious problem in relation to immediate forming, the H from that source seldom results in permanent injury, and prompt heating after pickling is a practically sure cure.

The solubility of H is much greater in austenite than in ferrite; so its removal is hindered by a "full" anneal.

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## CHAPTER 10

### PRECIPITATION HARDENING

Another treatment which may be applied to certain steels, either ferritic or austenitic, without producing transformation in the steel matrix itself, is precipitation hardening. It is often considered that low-C steels are not affected by quenching and tempering, which is true in the sense that such steels demand an extremely high rate of cooling to produce martensite, so that, unless the steel is in very small section and the quenching very drastic, an attempt to quench results only in the usual normalized, rather than the hardened, structure.

**Behavior of Low-Carbon Steel.** If, however, a very low-C steel, say 0.05% C, is quenched, not from above  $A_{c3}$ , say 1700°, as usual, but from below it, about 1300°, and then tempered for perhaps 10 hr at 140°, the Brinell hardness as quenched may be 120 and as tempered 150. If we let the steel merely stand at room temperature, it gets harder, in a month it may reach 170 Brinell, and its elongation may drop from an original 25% to 15%. A similar but less marked change is noted in higher-C steels, quite discernible at around 0.15% C; but in still higher-C steels the phenomenon is less readily observed because of the higher hardness of such steels.

Briggs,<sup>1</sup> however, found that an ordinary cast steel of 0.25 C, 0.34 Si, 0.70% Mn quenched from 1250° and aged for 2 hr at 212°, 20 hr at 120°, or allowed to rest at room temperature for 30 days, increased in Brinell hardness from 155–160 annealed, and 160–165 as quenched to 175 for the elevated-temperature aging and 190 for the room-temperature aging. The aging increases tensile and yield strengths with decrease in ductility. The properties of this steel annealed at 1650° with and without subsequent water quenching from 1250° and 20 hr reheating at 120° are shown in Table 17.

The hardness of the material aged at 120° did not change after 6 months' storage at room temperature. In discussion, Herty stated that rolled steel of the same composition did not age-harden under the same treatment and ascribed the difference to the microsegrega-



TABLE 17

	Annealed Only	Annealed, Quenched, and Aged
Tensile	79,000	85,500
Yield	55,000	63,000
Elong., %	25.5	19.5
R.A., %	40	33

tion of C in the grains of the cast steel as against more uniform distribution of C in the wrought steel.

Briggs' tests were made on 0.505-in.-diameter bars. The indications are that a rapid rate of cooling is necessary in the quench to put the steel in condition for precipitation hardening, so that the treatment would probably not be effective on large castings with heavy sections.

**Mechanism of Precipitation Hardening.** From such observations as these it is deduced that the effect is due to solution of cementite

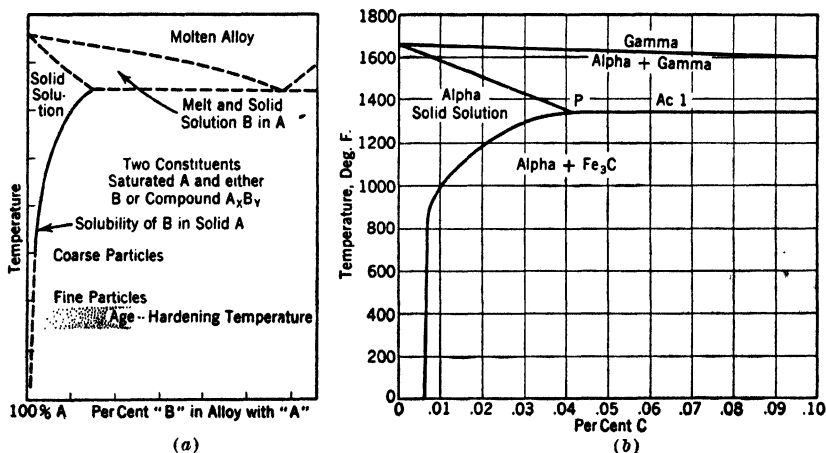


FIG. 135. (a) Typical constitution diagram illustrating the solid-solubility characteristics necessary for age hardening. Schematic diagram. (Davenport and Bain<sup>2</sup>). (b) Solubility of cementite in ferrite. This is an enlarged plot of the same field in Fig. 3.

in ferrite at 1200–1300° and precipitation of cementite from ferrite at lower temperatures. Further study shows that the solubility curve for cementite in ferrite has the shape shown in Fig. 135b. Its absolute location is still somewhat in doubt.

This has the form of the typical "precipitation-hardening" curves, schematically shown in Fig. 135a, and familiar to nonferrous metal-

lurgists, especially in respect to duralumin and to copper-beryllium alloys. The behavior of the quenched material on tempering at different temperatures and times is like that of precipitation-hardening alloys in general, the hardening taking place so slowly that the process is usually called "aging" or age hardening rather than tempering. The term tempering, which usually connotes some degree of softening, is not suitable, because in these cases tempering does not progressively soften the material.

The results of quenching and reheating in a precipitation-hardening alloy are just reversed from those of martensitic quenching. In

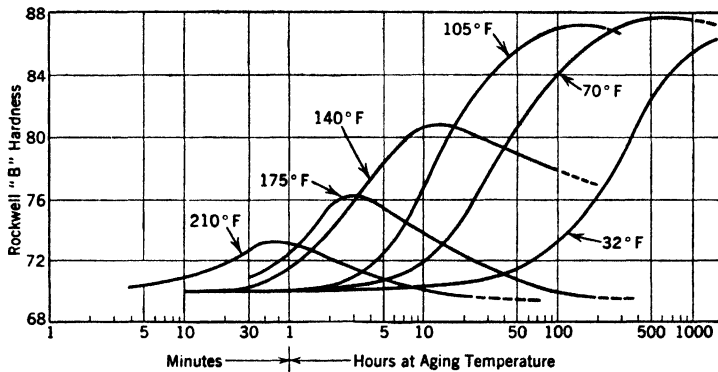


FIG. 136. Quench aging of 0.06% C steel after quenching from 1325°, plotted on logarithmic time scale. (Davenport and Bain<sup>2</sup>)

the case of ferrite and cementite, like other precipitation-hardening alloys, quenching *softens* and *toughens*, whereas reheating to a lower temperature *hardens*.

However, long-continued reheating will again soften the material, as Fig. 136 shows.

**Solution and Separation.** The explanation for these phenomena is that at the quenching temperature, above the solubility line, cementite is taken into solution and retained in solution on quenching. In the cold state the ferrite cannot firmly hold the cementite in solution; the cementite wants to precipitate out. If the steel is warmed somewhat, or even on very long standing, the cementite does come out, in very tiny particles. At room temperature these tiny particles remain tiny and distributed, but at slightly higher temperatures they diffuse, agglomerate, and spheroidize, just like the more notable spheroidization in a spheroidizing anneal of pearlite. In the agglomerated condition these larger particles of cementite do

not have so much hardening effect as when they are smaller and more uniformly distributed. This effect is shown in Fig. 136.

Such phenomena occur within the ferrite of ordinary steels on cooling from temperatures above the cementite solubility line. If the steel is not quenched from above the solubility line, but slow-cooled, the carbide precipitation may occur in grain boundaries and cleavage planes and embrittle the steel, whereas, if it is quenched and then allowed to stand, or only slightly reheated, the carbide will precipitate throughout the grains, and embrittlement will not ensue; merely a hardening and strengthening will occur.

An entirely similar situation exists with N. Although the normal N content of steel is only a few thousandths of a per cent, even this small amount can have an effect. Nitrogen may be present in the steel, as cast, as molecular  $N_2$ , in solid solution, or it may be combined as nitrides of Al, Ti, Zr, or other elements. Any N not combined as a nitride stable at high temperature, can produce Fe nitride, perhaps  $Fe_3N$ , on cooling, and the nitride can be redissolved on heating.

Oxygen present as the gas, or as iron oxide in solution, not combined as stable oxides of Al, Ti, etc., is thought to behave similarly. Hence a solution treatment by heating, followed by cooling and aging, may produce effects of the precipitation-hardening type in which any or all of the elements C, N, and O are concerned. It is seldom clear which one is the chief culprit. The preponderance of evidence indicates that both C and N produce precipitation-hardening embrittlement, with the effect of C most easily distinguished in low-C steels.

**Temper Brittleness.** The "quench-aging" phenomenon is not very much of a help or very much of a nuisance in ordinary C steels with Mn below 0.70%. In certain alloy steels, notably those containing considerable Mn, say 1.25–2%, and those with commercial amounts of Ni and Cr, it is found that, after the regular martensitic quenching and tempering, with tempering temperatures around 800–1100°, for instance, the steels are brittle in impact if slowly cooled from the tempering temperature, but tough if quenched from that temperature. This is termed "temper brittleness" or, in German, "Krupp Krankheit," that is, "Krupp sickness," since the Ni-Cr steels in which it is prevalent were early known as Krupp compositions.

Something is separated out. The something can be carbide or nitride. If it be postulated that it is carbide, we remember that cementite is not pure  $Fe_3C$  in steels alloyed with carbide-forming

elements, like Mn, Cr, Mo, V, but instead is a more complex carbide. The complex carbides containing considerable Mn or Cr seem peculiarly prone to grain-boundary precipitation on slow cooling of ferrite in which they are dissolved, and of course the presence of other metals dissolved in ferrite affects the ability of the ferrite to dissolve and hold or to precipitate out carbides or nitrides.

Cementite carrying Mo appears to diffuse less readily and not to be precipitated at the grain boundaries. In low-alloy steels and small sections, 0.20% Mo ordinarily removes temper brittleness. With some steels in large sections, 0.50% may be required, and even this may only mitigate and not avoid the phenomenon.

Although the exact solubility curves for carbides and nitrides in these temper-brittle alloy steels are not known, it is plain that a precipitation-hardening effect is involved, probably accentuated by a smaller solubility produced by the introduction of the alloying element.

**Temper Brittleness Not Revealed by Static Tests.** Whereas temper-brittle steels show normal ductility in the tension test, they are poor in notched-bar impact and considered undesirable for shock-resisting service. The magnitude of this effect is rather large. A temper-brittle 0.30 C, 3.50 Ni, 1% Cr steel may show 5 ft-lb Izod on furnace cooling after tempering and 50 if water-quenched from the tempering temperature.

The ratio of room temperature Izod or Charpy values between specimens slow-cooled and those quenched after tempering is called the "susceptibility ratio," but such a ratio is an inadequate criterion, since the type of notch, the specimen dimensions, and the testing temperature all alter the values. Notched-bar impact values over a range of low temperature are more revealing, and the temperature at which brittle failure begins is a better index.<sup>80-81</sup>

Reheating of a quenched nonembrittled steel into the 800°-1100° tempering range will embrittle it. The phenomenon is a reversible one.

**Slack-Quenched Steels.** The less tough the quenched unembrittled steel is, the more prone it is to show temper brittleness. Transverse specimens may reveal it more clearly than longitudinal ones. Steels slack-quenched in hardening, so that they contain ferrite or bainite along with martensite are more prone to temper brittleness than is a tempered structure derived from martensite alone. Little is known about the effect of cold work on temper brittleness. It might be an added factor in the case of the high-Mn high-S cold-

drawn steels discussed on p. 241. Cold work usually facilitates precipitation hardening.

Hollomon<sup>8</sup> has given an extensive résumé of available information on the phenomena and theories of temper brittleness.

Not all heats of compositions that are normally temper-brittle turn out to be temper-brittle, and the range of tempering temperatures that produces temper brittleness may differ from heat to heat, but rapid cooling after tempering is a wise precaution when this class of steels is being handled. If the section treated is so large that the interior cannot be cooled rapidly by quenching, an alloying element should be added that will minimize temper brittleness. Fortunately Mo has this effect. If a part of the Mn or Cr of temper-brittle steels is replaced by Mo, temper brittleness is decreased.

**Size of Particles.** In all this discussion of precipitated carbides and other precipitation-hardening materials it should not be thought that the carbides or other precipitated materials are necessarily visible under the microscope. As a rule they are not visible till they have been agglomerated and spheroidized to such a size that the matrix ceases to be appreciably hardened by their presence, but the existence of submicroscopic hardening particles is thoroughly accepted by metallurgists on the basis of their effects, even when invisible.

**Strain Aging.** In precipitation hardening in general, it is found that cast material responds less readily to solution and aging treatments than work-hardened, that is, strained and deformed material.\* This effect crops up in cold-rolled low-C steels. These rimmed un-killed steels, as usually made, contain both C and O, as well as small amounts of N. Bessemer steels are especially high in N.

Now O and N or, more strictly, Fe and Mn oxides and Fe nitride show solubility curves in ferrite of the same general character as that of Fig. 135*b* for cementite in ferrite.†

In quench aging it seems likely that carbides are at least partly responsible. But aging after cold rolling, that is, strain aging, appears to be more definitely due to precipitation of oxides and nitrides than to carbides, though all may be concerned. Weber and coworkers<sup>20</sup> ascribe it to carbides plus nitrides, Altenburger<sup>22</sup> to nitrides.

\* An exception noted in the work of Briggs and of Herty on quench aging of cast and wrought 0.25% C steel, mentioned at the beginning of this chapter, does not affect the validity of the general rule.

† The solubility curve for Fe nitride, according to Séferian,<sup>7</sup> shows dissimilarity at a N content above that present in the steels under discussion.

**Separation During Testing.** The assumed situation in strain aging is that the steel has in solution, not previously precipitated, some compound that wants to come out. During deformation, as in cold rolling or in compression or tensile testing, it starts to come out, and it may continue to be precipitated thereafter at room temperature, or more quickly on slight reheating. Such compounds so pre-

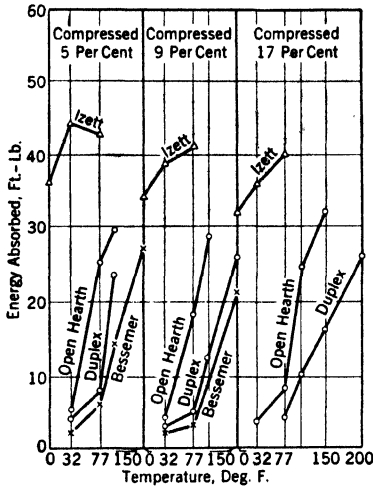


FIG. 137.

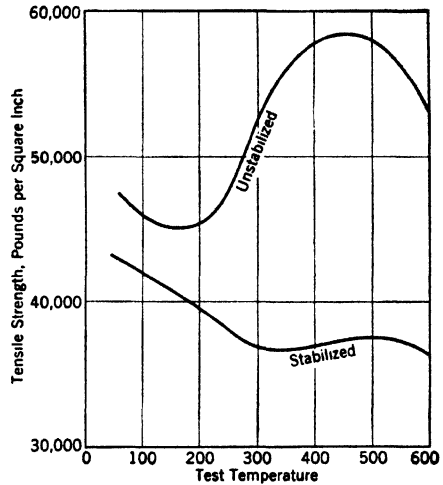


FIG. 138.

FIG. 137. Effect of lowered temperature on impact resistance (Charpy bar, key-hole notch) after various degrees of cold work by compression. The Izett steel was treated with Al, which is considered to combine with O and N to form insoluble compounds, while the other steels contain soluble ones. The sensitivity to lowered temperature increases with the amount of soluble O and N. (Epstein <sup>4</sup>)

FIG. 138. Tensile strength at elevated temperature of ordinary low-C steel (unstabilized) and of similar steel stabilized by addition of Al and Ti. (Kenyon and Burns <sup>5</sup>)

cipitated tend to come out in the grain boundaries and cleavage planes and make the steel more brittle. The impact resistance of cold-worked and aged steels, compressed and allowed to age, is shown in Fig. 137 from Epstein.<sup>4</sup>

The precipitation effect is so great in ordinary mild steel that it occurs while a tensile bar is being pulled at slightly elevated temperatures, as is shown by the peak in the tensile strength of the mild steel in Fig. 138.

Cold-worked low-C steel tested immediately after being slightly cold-worked has a smooth tensile stress-strain diagram.<sup>5</sup> If allowed to age it shows a jogged one, Fig. 139.

## PRECIPITATION HARDENING

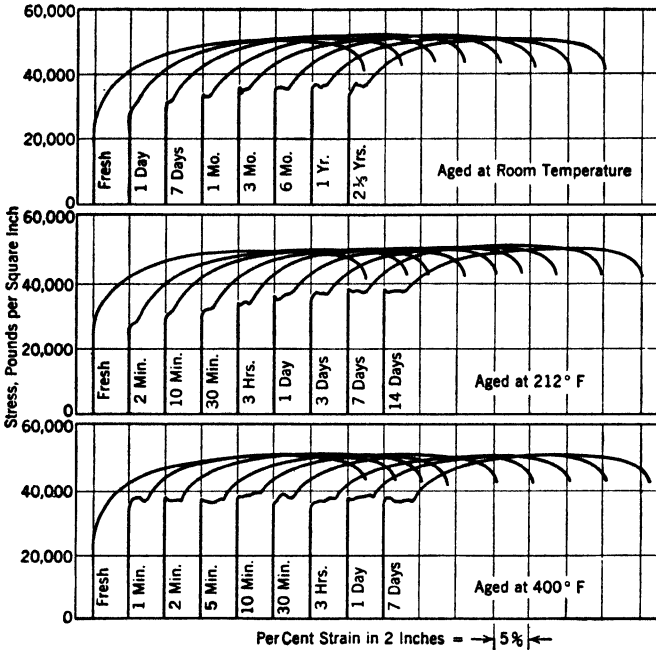


FIG. 139. Stress-strain curves for mild steel sheets cold-rolled 1% and aged for the times and at the temperatures shown. The steels that show a smooth curve will not stretcher-strain; those that show a jugged curve will. The jog at the yield point returns more rapidly at increased aging temperature. (Griffis, Kenyon and Burns <sup>6</sup>)

**Stretcher Straining.** The presence of this jog leads to a difficulty in deep drawing known as “stretcher straining.” Davenport and Bain <sup>2</sup> have cited evidence which shows that stretcher straining is due to a precipitation-hardening action.

The effect of temper rolling tends to “wear off” in time. Under one conception this is due to stress relief (which is rapidly brought about by moderate heating) occurring slowly at room temperature. Ott <sup>8</sup> reports that auto body stock prepared in two ways, (a) cold-reduced, annealed, temper-rolled 1%, (b) hot-rolled, normalized, annealed, temper-rolled 1%, behaved as shown in Table 18. The other conception is that such behavior indicates precipitation hardening.

**Stabilization.** As Griffis and coworkers <sup>6</sup> have shown, addition of 0.04% Al + 0.06% Ti to a 0.05% C steel stabilizes the steel and prevents the action, whereas, as Epstein <sup>4</sup> has brought out, the Al

TABLE 18

<i>Type</i>	<i>Time</i>	<i>Tensile Strength, Psi</i>	<i>Yield Strength, Psi</i>	<i>Elonga- tion, %</i>
(a)	Immediately after temper rolling	45,000	24,000	42
	3 months later	45,000	32,000	39
(b)	Immediately after temper rolling	50,000	28,000	40
	3 months later	50,500	35,000	38

addition to "Izett" steel also prevents strain aging. Since Al and Ti can readily combine with O and N to form inert oxides and nitrides that will not readily go in and out of solution, this effect makes it pretty clear that in the "unstabilized" steels it is a precipitation-hardening effect that is concerned. Large additions of Ti to low-C steel produce a very stable titanium carbide, so stable that the steel behaves like carbonless ferrite containing inert inclusions of titanium carbide, rather than as the ferrite and cementite of ordinary steel.

**Blue Brittleness.** Many C steels which do not contain stabilizing elements are found to show a peak of strength and hardness and valleys in ductility and notched-bar impact values (not necessarily coincident) within the range 400–800°, when subjected to short-time high-temperature tensile tests on high-temperature impact tests. A blue "temper color" appears from surface oxidation at some temperatures in this range, when the tests are made in air; hence this phenomenon has been termed "blue brittleness."

The temperature of occurrence varies with C and with previous cold work. Obviously a steel exhibiting this phenomenon has reduced formability in the temperature range where it occurs. Removal of blue brittleness by addition of elements that combine with O and N indicates that it is another manifestation of precipitation hardening. The topic has been discussed by Sisco.<sup>9</sup>

**Utility of Precipitation Hardening.** These cases of precipitation hardening produce phenomena that one generally seeks to avoid, and they are commercially advantageous only in rather special cases. Carbide-precipitation effects are of great moment in some of the high-Cr steels that are welded or that are employed in high-temperature service. Not only the ferritic steels, but also the austenitic ones, like high-Mn steel and the familiar 18 Cr 8% Ni stainless steel, are subject to them. Such phenomena are mentioned in Vol. III. In the austenitic steels the precipitable material is only taken



into solution at a high temperature, usually 1800° or above, and precipitation may occur quite rapidly during cooling. It is therefore customary to quench heavy sections whose interiors would cool too slowly in air to avoid precipitation.

**Quench Annealing.** Because of this, a peculiar terminology has grown up, by which the high-temperature solution treatment and subsequent quenching are together termed a "quench anneal." Instead of the hardening sought in martensitic quenching, the quench anneal seeks exactly the reverse, that is, to preserve the softness of the austenite and *prevent* precipitation hardening. However, there are cases where precipitation hardening can be usefully applied to an alloy steel. There are many highly interesting cases of precipitation-hardenable ferrous alloys, most of which are too highly alloyed to come into the scope of this Volume, but a case, involving small amounts of a cheap alloying element, Cu, is of interest to the general heat treater.

**Copper Steels and Their Precipitation-Hardening Heat Treatment.** Copper forms no carbides in steel and enters solely into the ferrite. Its behavior in improving the atmospheric corrosion resistance of steel and its strengthening effect on ferrite are discussed in Vol. III. Copper increases hardenability in martensitic quenching. Its effect is analogous to that of Ni in this respect. But, unlike Ni, if present in sufficient amount, it can be made to strengthen steel through a precipitation-hardening heat treatment.

The solid solubility curve of copper in ferrite<sup>11</sup> is of the same type as that for cementite in ferrite, but the solubility is much greater. The exact location of the curve and the effect of other alloying elements on its position are not fully known, but in general any ferritic steel containing from perhaps 0.50 and certainly 0.70 to around 1.50% or more Cu is amenable to precipitation hardening, as shown in Fig. 140.

If such a steel is heated to 1475–1550°, that is, above the solubility line, the Cu will go into solid solution; and, if the steel is then cooled, the Cu remains in solid solution. Upon reheating, say to 930° for 4 hr, the Cu (itself carrying what little Fe it can hold) is precipitated. The hardness and strength are increased. The changes in hardness with different reheating times and temperatures are indicated in Fig. 141.

This phenomenon was first noted by Bird, in the tempering of quenched Cu steels. The usual gradual softening with increase in tempering temperature occurred up to 800° and again above 1100°,

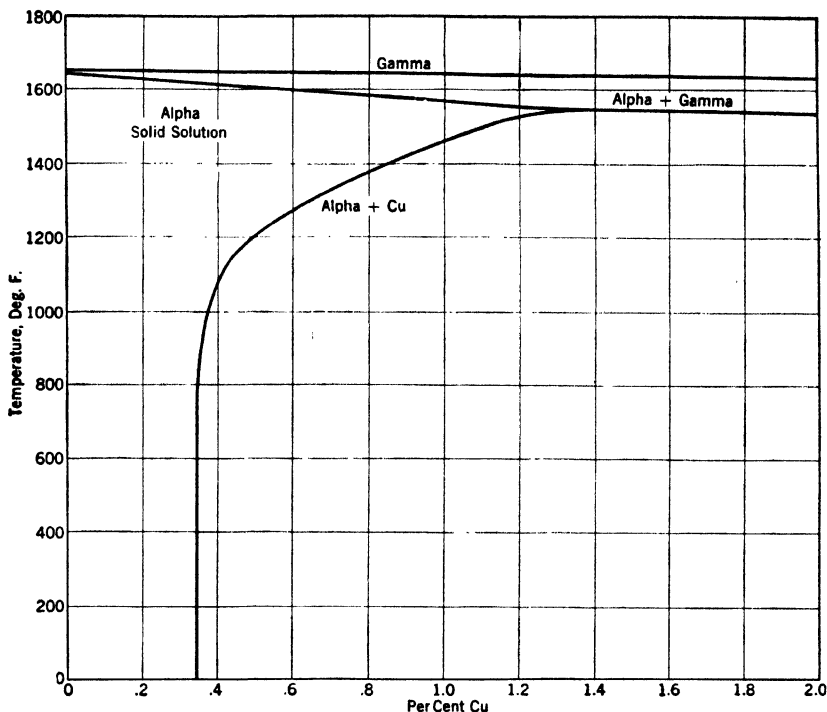


FIG. 140. Solubility of copper in ferrite.

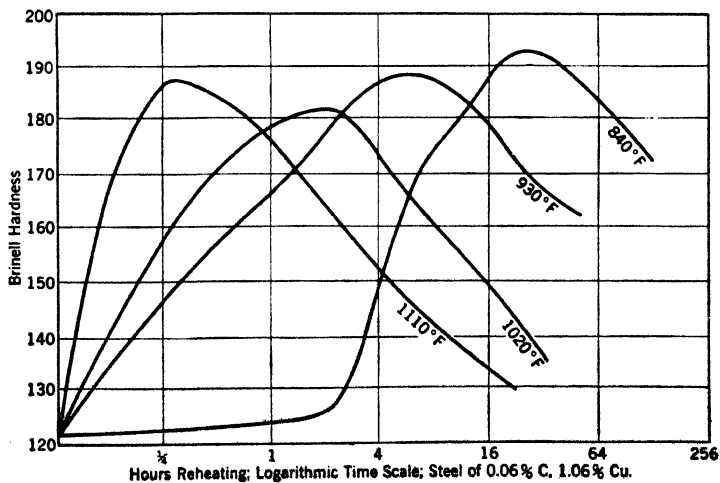


FIG. 141. Hardness of a Cu steel, quenched from 1500°, reheated for the times and at the temperatures shown. (Smith and Palmer<sup>10</sup>)

but in this interval the strength rose and ductility slightly decreased. Bird's curves are given by Gregg and Daniloff.<sup>12</sup>

Kinnear<sup>13</sup> first applied it to cast steel. Low- and medium-C Cu steels have their tensile and yield strengths increased by about 20,000 psi by such treatment, the ductility decreasing at the same time, but less than when the increased strength is obtained in a normalized steel by increasing the C.

**Large Sections.** The important feature from the point of view of heat treatment, is that the precipitation of Cu from solution is sluggish so that no severe quenching is required to hold it in solution ready for precipitation. A cooling rate as low as 1° per minute will retain it, so that even very large normalized sections are equally capable of throwing out the Cu at the center and at the edge upon reheating. Hence there is almost no mass effect with which to contend. Moreover, the precipitation goes on uniformly and gradually so that there is little tendency for distortion on reheating. Means by which the centers of heavy sections can have their strength boosted 20,000 psi are rare, so that the potentialities of precipitation-hardened Cu steels are great.

It is not necessary to allow the material to get cold and then to reheat it. The steel can be normalized so as to produce fine pearlite and when it has cooled a little below  $A_{r1}$  can be transferred to a furnace at, say, 930° and held there, when the precipitation of Cu will occur just as well as though the steel had been cooled and reheated to 930°.<sup>14, 15</sup>

Copper, like most alloying elements that enter ferrite, increases strength in non-heat-treated steels, and it does not interfere with the usual heat treatment by quenching and tempering. This effect, in steels not intentionally precipitation-hardened, is discussed in Vol. III. Of course, the tempering, if below the solubility line of Fig. 140, involves some precipitation hardening; and the tempering time can be adjusted so as to oppose the sorbitic softening of tempering by some Cu precipitation hardening.

**Effect of Composition.** The precipitation-hardening effect is not marked in eutectoid steel and is most effective in the lower-C ranges. The effect begins to be evident at about 0.70% Cu and is nearly as marked with 1% as with 2%. About 1.25% Cu is perhaps the optimum content when precipitation hardening is to be used.

The presence of other alloying elements, such as Ni, Cr, V, Mn, and Si, does not interfere with precipitation hardening by Cu.

Hence a steel can be alloyed to produce a given set of properties on normalizing, and the precipitation-hardening effect of Cu can be superimposed on those properties. Since this hardening involves some loss of toughness, it is advisable to reduce the C below that which would be used in its absence.

**Properties Obtained.** In order to bring out the improvement in properties due to precipitation hardening, data on steel castings in regular production, from Kinnear,<sup>18</sup> Zeuge,\* and Finlayson † are given in Table 19.

TABLE 19

	Per Cent					Tensile	Yield	Elong., %	R.A., %	Brinell
	C	Mn	Si	Cu						
Kinnear	0.29	....	....	0.94	N <sub>1</sub>	88,000	58,000	19	31	179
					R <sub>1</sub>	104,000	75,500	20.5	36.5	207
Zeuge	0.26	0.73	0.34	0.97	Re	106,000	76,000	21.5	43.5	
Kinnear	0.13	0.63	0.30	1.00	N <sub>2</sub>	70,000	50,000	31	54	143
					R <sub>2</sub>	89,000	70,000	23.5	49.5	196
Finlayson	0.11	1.04	1.23	1.74	A	81,000	64,000	31	56	
					ANR	106,500	88,500	25	53	

N<sub>1</sub>—Normalized from 1550°. R<sub>1</sub>—Normalized from 1550°, reheated to 950°.  
 N<sub>2</sub>—Normalized from 1600°. R<sub>2</sub>—Normalized from 1600° and reheated to 950°.  
 Re—Normalized from 1650°, reheated to 1000°F, held 4 hr.  
 A—Annealed at 1740°—slow-cooled. ANR—Annealed at 1740°, slow-cooled—reheated to, and air-cooled from, 1580°, reheated at 915°.

**Oxidation of Copper Steels.** The presence of Cu sufficient for precipitation hardening involves one drawback. In the soaking pit or in heating prior to rolling or forging, the surface of steel containing, say, 1% Cu, is selectively oxidized, iron oxide forming more rapidly than copper oxide so that a film of Cu appears beneath the scale. If the steel is heated above the melting point of Cu, the Cu tends to penetrate between the grains of the steel and produce a rough rolled or forged surface. Hence great care must be taken in controlling the heating temperatures. If a little Ni, say, a third as much Ni as Cu, is present, this intergranular penetration is avoided, so high-Cu wrought steels often contain some added Ni for this reason. Castings, which are not subjected to hot working, do not need the Ni-addition.

\*Sivyer Steel Casting Co.

† Private Communication, April 16, 1934.

**Ford Steels.** Copper is used in several "steels" developed by the Ford Motor Company. It uses Cu<sup>17</sup> in so-called cast "steel" for crankshafts, brake drums, and pistons. These are cast from alloys of about 1.50 C, 1 Si, 0.80% Mn, and from 1.50 to 3.00% Cu, with the highest Cu only in the brake drums. The brake drums also carry about 0.15% Mo, the crankshaft about 0.45% Cr, and the pistons about 0.10% Cr. These alloys are cast white, like malleable iron, and partially malleablized to give temper C and a pearlitic matrix. The castings are then normalized or annealed. The details for the crankshaft are: C 1.35–1.60, Cu 1.50–2.00, Si 0.85–1.10, Mn 0.60–0.80, Cr 0.40–0.50—P 0.10 max., S 0.06% max. Heat to 1650°, hold 20 min, air-cool to 1200°, reheat to 1400°, hold 1 hr, furnace-cool in 1 hr to 1000°. Properties obtained are 107,500 tensile, 92,000 yield, 2.75% elong., 2.25% R.A., 270 Brinell. The presence of Cu in these alloys is primarily to hasten graphitization in the short anneal for malleablizing and to add fluidity in casting.<sup>18</sup>

These high-C alloys are not precipitation-hardened, since it would add little strength to the matrix, which is practically of eutectoid composition and hence too high in C to respond much. Work by Russell<sup>19</sup> has, however, indicated possible advantage in increased ability to withstand overstress in fatigue by carrying out the precipitation-hardening operation, even though the static properties and the endurance limit are not improved.

**Other Precipitation-Hardening Steels.** In nitriding steels (see Vol. III) one may desire to back up the hard nitrided surface by a strong core. It would be convenient if the steel could be machined in soft condition before nitriding and hardened by precipitation hardening during nitriding. By utilizing the fact that Ni–Al steels may be precipitation hardened, this can be accomplished. Most of the nitriding steels contain Al anyhow. To such, the addition of Ni, as French and Homerberg<sup>20</sup> have shown, allows precipitation hardening to occur during nitriding.

A steel of 0.20 C, 0.50 Mn, 0.25 Si, 1.0 Al, 1.1 Cr, 0.25 Mo, and 3.50% Ni, after nitriding 48 hr at 975°, has a yield strength of 115,000, whereas without the Ni and the precipitation-hardening treatment the yield strength of the core would be around 85,000. The hardness of the nitrided case is somewhat decreased by the presence of Ni, but its toughness appears somewhat improved. Ward<sup>21</sup> reported even higher properties for a steel of this class in nitrided parts for aircraft motors.

That the precipitation hardening is due to some compound of Ni and Al has been shown by Foley,<sup>22</sup> who produced it in a Ni-Cr steel to which Al was added. He found that elimination of the Cr did not affect the precipitation hardening, but leaving out either Ni or Al eliminated it.

Foley made a steel of 0.03 C, 0.16 Mn, trace of Si, 5.03 Ni, and 2.59% Al. By quenching from 1300° or above, the Ni-Al compound is held in solution. On reheating, as shown by Fig. 142, the Brinell hardness could be doubled. The reheating curves are of the type

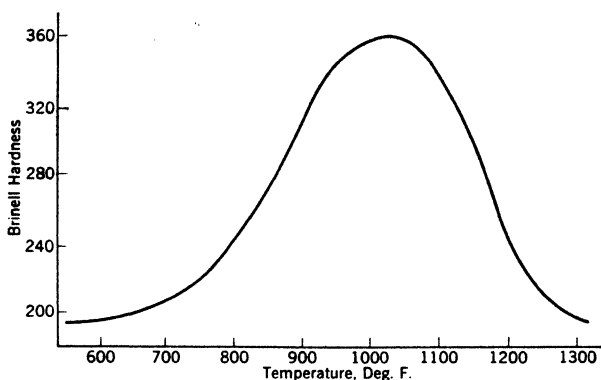


FIG. 142. Change in hardness on reheating for 12 hr, at temperatures shown, a steel of 0.03 C, 2.60 Al, 5.00% Ni, after water-quenching from 1300°. (Foley<sup>22</sup>)

characteristic of precipitation hardening; 800° does not produce full hardness in a very long time, 1000° requires 24 hr, 1050° makes it fully hard in 4 hr without decrease in hardness on longer holding, whereas 1100° also makes it fully hard in 4 hr, but further heating produces softening. Thus the 48-hr 975° treatment used by French and Homerberg is entirely consistent with Foley's results.

That the same optimum precipitation-hardening temperature of 950–1050° holds for the Ni-Al combination in the presence of other elements is shown by Foley's experiments on a valve steel of 0.50 C, 0.36 Mn, 0.94 Si, 1.58 Ni, 10.66 Cr, and 1.77% Al. The tensile strength of this steel was raised from 130,000 in the quenched state to 180,000 by reheating at 1000°.

In Ni-base alloys, Si, Al or Ti make precipitation hardening possible. Compounds are formed. Such compounds of Ni can be produced in some Cu-base or other alloys.

By adjusting the composition of an 18:8 type of austenite stainless steel (by reducing the percentages of elements favoring stabil-

ity of austenite and increasing those favoring transformation to ferrite), so that on cooling transformation of at least a considerable part of the austenite to ferrite occurs, a partially ferritic steel is produced. The ferrite can be precipitation-hardened by separation of some compound. Thus, it is stated<sup>23</sup> that a steel of 0.07 C, 17 Cr, 7 Ni, 0.50 Mn, 0.85 Ti, 0.15% Al can be materially strengthened, to over 200,000 psi tensile by precipitation hardening without resorting to cold work. The exact precipitation-hardening compound concerned is not certain. Columbium can be used instead of Ti.

**Secondary Hardening.** So-called "secondary hardening" of quenched and tempered steel, by which, instead of becoming softer with increase of tempering temperature, the steel hardens or at least maintains its hardness, is ascribable to precipitation hardening. This phenomenon is important in high-speed steels containing V (compare Fig. 209).

The strength of a 0.30 C, 2.60 Ni, 0.75 Cr, 0.50% Mo steel, oil-quenched from 1650°, falls as the tempering temperature rises, whereas if 0.30% V is present the tensile strength remains practically constant to 1100°, and the yield strength rises. Abram<sup>24</sup> shows the data given in Table 20.

TABLE 20

	No V		With V	
	Tempered		Tempered	
	850°	1100°	850°	1100°
Tensile	197,500	170,000	196,000	199,000
Yield	176,000	154,500	184,500	190,000
Elong., %	14	18	14	16
R.A., %	49	59	52	55
Izod	22	42	20	23

Abram ascribes this to the solution of vanadium carbide in austenite on the quench, its retention in solution in the ferrite formed from austenite on quenching, and its precipitation on reheating.

**Cobalt, Molybdenum, and Tungsten in Precipitation-Hardening Alloys.** Various compounds that can be taken into and thrown out of solution will produce precipitation hardening. Among these are compounds of Fe with Mo or W. Carbonless alloys of Fe with  $\beta$  to nearly 40% Mo are precipitation-hardenable. One of 30% Mo quenched from 2600° and reheated 20 hr at 1155° rose from 40 Rockwell C to 65. The compound  $Fe_3Mo_2$ , or a phase (epsilon) of about that composition, is precipitated, according to Sykes.<sup>25</sup>

Sykes<sup>26</sup> has also shown that carbonless alloys of 8 to 33% W are similarly precipitation hardening through the separation, on reheating, of  $Fe_3W_2$  or the epsilon phase of that series. A 25% W alloy, quenched from 2730° and aged as shown in Fig. 143, gave a Brinell hardness of 185 to 370.

Complex steels in which large amounts of Mo or W appear, therefore, have possibilities of precipitation hardening by quenching to

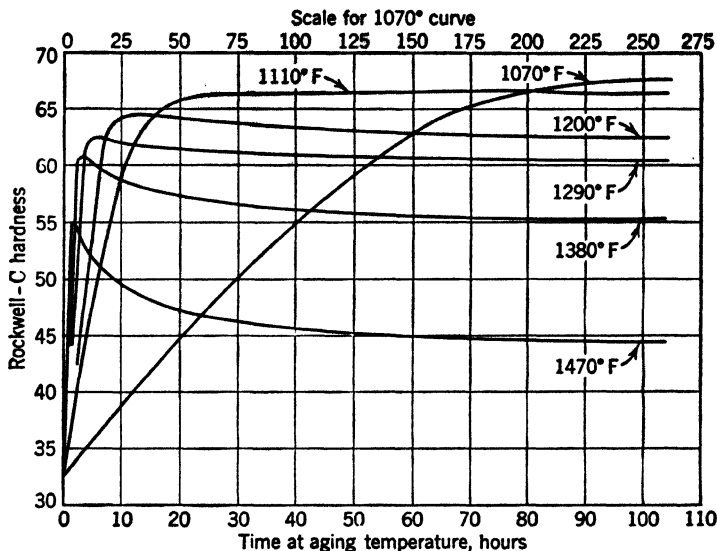


Fig. 143. Typical age-hardening or quench-hardening behavior. Hardening of an alloy of 75 Fe, 25% Mo, after quenching from 2600° and reheating as shown. (after Sykes<sup>25</sup>)

produce solution and reheating to produce precipitation of molybdenides or tungstides. Owing to the hardness and stability of such compounds, alloys containing them offer promise as tools and dies for hot working. Cobalt, already used as a bond and matrix for sintered-carbide tools, appears to serve as a solvent for such compounds.

Sykes<sup>27</sup> discussed the precipitation-hardening relations of Fe-Co-W alloys. On the basis of those phenomena and related ones in Fe-Co-Mo alloys, still more complex precipitation-hardenable alloys are being built up.

Harrington<sup>28</sup> describes an alloy of 1 C, 0.4 V, 6 Cr, 8 Mo, and 36% Co, remainder Fe, which, oil-quenched from 1740°, shows 42 Rockwell C. Such an alloy was expected to be suitable for cutting tools of a general class between high-speed steel and sintered-



carbide tools. With the Cr, V and C contents this alloy probably owes some of its hardness to carbides as well as molybdenides.

Green \* secures a similar range of hardness in an alloy of 0.12 C, 2 V, 3 Mo, 19 W, 30 Co, 0.20 Si, 0.30% Mn, remainder Fe, which may owe its hardening power primarily to the tungstide rather than to carbide. In its treatment the forged stock is heated 5 hr at 1625° and slow-cooled, heated to 2250° for a very short time, quenched in lead at 1100°, and then air-cooled. This serves as the solution treatment and produces a machinable material of about 46 Rockwell C. Reheating to 800° for an hour hardens to 56 Rockwell C, to 975° for 20 min to 66 Rockwell C, and to 1100°, for a still shorter time, to 71 Rockwell C. The exact treatment for best life of tools has still to be decided. Complex heat-resistant alloys using this group of elements and others with precipitation-hardening potentialities, are receiving a great amount of attention.

Permanent-magnet alloys of the Al-Ni-Co groups, in which precipitation hardening is applied, are phenomenal in magnetic properties (see Vol. III).

These examples will serve to indicate that whereas quench hardening through the formation of martensite can be applied to a much wider range of steels, the possibilities of precipitation hardening in those steels to which it applies are also of decided interest. The heat treater today needs to understand the principles of precipitation hardening as well as the principles of quench hardening.

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## SECTION IV. HEAT TREATMENT INVOLVING TRANSFORMATION OF AUSTENITE

### CHAPTER 11

#### HARDENABILITY

The thermal and dilatometric curves of Figs. 6-11, Chapter 2, showed that, at a sufficiently rapid cooling rate, the transformation of austenite to ferrite and pearlite, or pearlite and cementite, can be suspended, in spite of the equilibrium diagram (Fig. 3), which might be thought to indicate that the transformation should be completed at 1335°. The diagram does *not* apply to other than equilibrium conditions.

The retention of austenite to lower temperatures and the adjustment of conditions so that it transforms at a *selected* lower temperature or temperature range are the essentials of heat treatment in its broadest sense. Section III, Chapters 9 and 10, dealt with effects in which austenite was not involved; heating was always below the critical temperature in process annealing and in precipitation hardening of ferrite and steels.

We now turn to heat treatment that does involve austenite. The production of austenite, by heating above the critical range, wipes out the initial structure of the cold steel, changing alpha to gamma iron, and, if sufficient time is given, producing a homogeneous austenite, with the C uniformly distributed throughout the gamma iron (Fig. 16). With sufficient time and temperature the austenite crystals, small when first formed, grow to larger size. The effect of this is retained even on slow cooling, as schematically indicated in Fig. 38.

The condition of the austenite prior to cooling greatly affects the behavior during cooling; inhomogeneous and homogeneous austenite, fine-grained and coarse-grained austenite behave very differently; hence all statements as to cooling behavior need scrutiny to establish just what kind of austenite is being discussed.

In the areas *GPS* and *ESQ* of Fig. 3 part of the austenite has to change its C content in order to remain stable at lower temperature. In the first area, Fe above its eutectoid concentration, in the second,  $\text{Fe}_3\text{C}$ , above its eutectoid concentration, is thrown out on cooling.

**Effects of Cooling.** All of these area boundaries are lowered on rapid cooling, except the  $1420^\circ$  beta change,  $\text{Ar}_2$ . The amount of lowering depends on the rate of cooling as well as on the condition

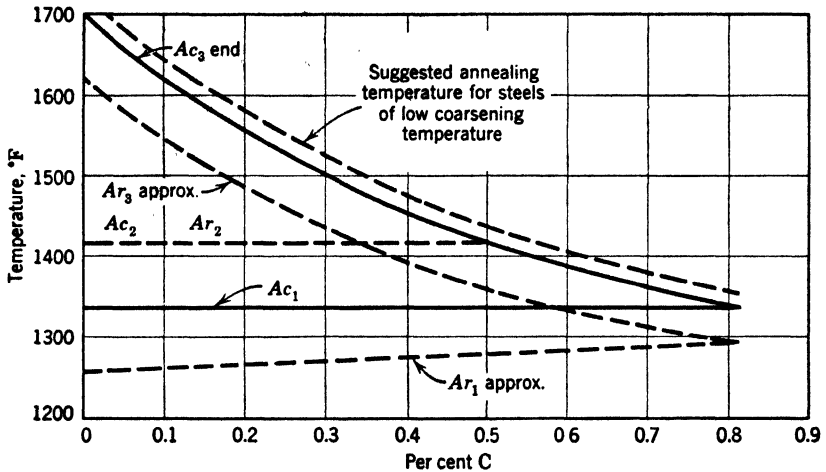


FIG. 144. Relation of critical range on heating and cooling to temperature for full annealing of hypoeutectoid-C steels.

of the austenite. But even on slow cooling, and with inhomogeneous austenite, depression of the boundaries occurs as indicated in Fig. 144.

Below  $1335^\circ$  in a C steel (or whatever temperature the line *PSK* of Fig. 3 comes at in an alloy steel), all of the austenite is unstable; it wants to transform. When it does transform (gamma iron changing to alpha or, in martensite, to a tetragonal crystal very similar to alpha), a new set of crystals is born. Many crystals appear where there was previously but one austenite crystal. As it is often put, transformation "refines the grain." The ferrite grain size itself depends on the temperature at which it is formed, and the distribution of carbide in the ferrite matrix likewise depends on that temperature.

In the cooling of austenite there are two domains, one extending from just below the critical to about  $600^\circ$  (or lower, depending on composition), a second below that temperature. In the upper domain, austenite transforms, depending on the C content, to ferrite

plus pearlite, to pearlite, or to pearlite plus cementite at the higher temperature or to acicular ferrite-plus-carbide structures (bainite) at the lower ones. In the second domain, from about 600° down, austenite transforms to martensite, or to lower bainite. The processes of annealing, normalizing, and "austempering" (and the transformation in the interior of shallow-hardening steels during quenching) fall into the first domain; martensitic quenching and "martempering" into the second. Operations in the first domain usually seek to use up all the austenite within that domain, preventing retention of any austenite.

In quenching, either in the usual coolants or in the "martempering" hot quench, the aim is to use no austenite in the upper domain, preserving it all for transformation to martensite.

When the aim is to use up austenite in the upper domain, cooling must be slow enough to give time for transformation and avoid retention. When the aim is to preserve austenite, cooling must be rapid.

**Trigger Action.** Two factors, inhomogeneity and grain size, complicate the situation in evaluating behavior of the steel, as it follows some single time-temperature cooling curve. First, there is a "trigger action" when the austenite is inhomogeneous.

As the cold steel, containing ferrite and carbide, in one or another structure, is heated past the critical, the alpha ferrite, originally nearly carbonless, changes (at a much lower temperature than it would change without pickup of C) to gamma austenite, by dissolving C from the cementite plates or spheroids which existed in the cold steel. The gamma which is formed right next to a piece of cementite is high in C, but that C still has to be passed on by diffusion to the gamma that is more remote. As the more remote gamma becomes richer in C, the concentration difference, which supplies diffusion head, analogous to hydraulic head, grows less. At anything short of complete diffusion and equalization of C content everywhere, the austenite grain remains richer in C than the average where it touched cementite, lower than the average at those locations most remote from cementite.

When austenite is cooled far enough so that it begins to revert, reversion starts at the boundaries of the austenite grains. Low-C austenite reverts more easily than that which carries the eutectoid amount of C, so that, if the austenite is not homogeneous, the low-C regions change over first, and, once the change is started, it tends to go on by "trigger action," so that the higher-C regions also tend to

change more quickly than they would by themselves, if not in contact with material poorer in C. Alloying elements, in general, diffuse much more slowly than C, and in alloy steels, homogeneity and the resultant behavior on cooling are dependent on the diffusion of the alloying elements much more than on that of the C. Differences in alloy concentration produce even more notable trigger action, one probable reason for the spread in the "hardenability bands" of alloy steels.

Austenite of higher-than-eutectoid-C content also reverts more easily than that of eutectoid-C content, and especially so if there are any undissolved carbides; for these have a similar "trigger action" (Fig. 7).

**Homogenization.** Hence, the homogeneity of the austenite controls the avidity for reversion on cooling. Fully homogeneous austenite is much more sluggish in reversion than inhomogeneous austenite.

Thus, the opportunity for diffusion given the austenite is an important factor in the cooling behavior.

Unless the steel has been provided with some "grain-growth inhibitor," after time and temperature ample to produce considerable homogenization, the austenite, originally fine-grained as it first passed from ferrite into austenite, begins to grow larger grains. The larger the grain (the less the total area of grain boundaries, where reversion starts), the more sluggish is reversion on cooling. The two factors, homogeneity of austenite and grain growth, tend to appear together, and so it is hard to untangle their individual effects. We are not easily able to appraise homogeneity, save by its effects during cooling, whereas we can measure the austenitic grain size; so discussion of the effects of austenitizing time and temperature is usually phrased in terms of grain size.

In hardening by quenching, the first hurdle to be passed occurs at  $A_{r1}$  and the temperature range immediately below it, that is, say,  $1300^{\circ}$ – $1000^{\circ}$ . The lower the temperature to which austenite is brought below the equilibrium temperature,  $1335^{\circ}$ , the greater its urge to revert. Its ability to revert is, however, limited by the increasing rigidity of the steel as it gets cooler, so that time required to start reversion decreases to a minimum, as the temperature drops to  $1100$ – $900^{\circ}$ , and then increases. *However, even at this minimum time there is a time delay before reversion starts.* The austenite, carrying C in solution, has to do some rearrangement of its atoms before it can throw out practically carbonless ferrite and  $Fe_3C$ . In the austenite, each C atom is surrounded by many more Fe atoms

than three, and the excess Fe must be elbowed out of the way, into a practically C-free aggregate, before ferrite can deposit. This elbowing time is called the "nucleation period," the "incubation period," or the "time delay."

**Ar' and Ar''.** If there were no delay at all, it would be impossible to cool rapidly enough to bring austenite down below this 1100–900°

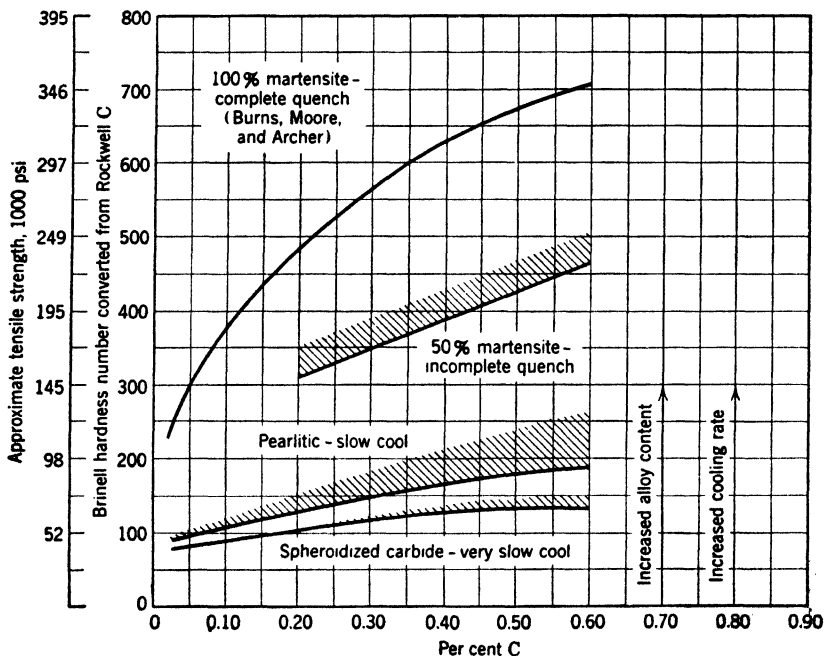


FIG. 145. Approximate tensile strength obtainable for steel in various conditions of heat treatment as a function of C content.

range and thus retain it for another transformation, that to martensite, for which the temperature must fall to 600° or below (depending on steel composition). The temperature at which *martensite starts* to form is designated as the  $M_s$  point, that at which the transformation is finished, as  $M_f$ . The transformation to ferrite plus pearlite or pearlite plus carbide,  $Ar_1$  (1335° on the equilibrium diagram) is, as we saw in Chapter 2, called  $Ar'$  when, by rapid cooling, it is brought down to 1100–900° range; and in the older terminology, the  $M_s$  point was called  $Ar''$ . At some intermediate rate of cooling, where part of the austenite reverts at  $Ar'$  and part is retained to revert at  $Ar''$ , this older terminology spoke of a "split transformation," as was shown in Figs. 6–11.

Griffiths and coworkers<sup>1</sup> make the suggestion that the notation  $Ar''$  be reserved for lower bainite, and the  $M_s$  to  $M_f$  range be termed  $Ar'''$ . A steel which does not give bainite would then be said to show  $Ar'$  and  $Ar'''$  rather than  $Ar'$  and  $Ar''$ . However, it is so difficult to untangle  $Ar''$  and  $Ar'''$  by thermal or dilatometric meth-

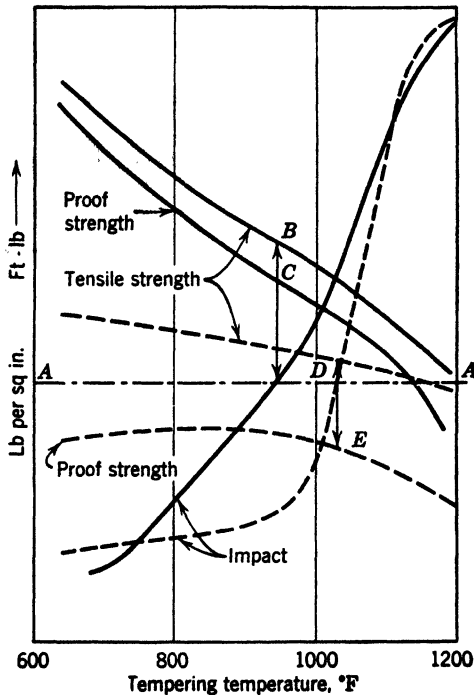


FIG. 146. Effect of tempering on steels transformed in the martensitic range (full lines) and in the intermediate range (dotted lines). (Griffiths, Pfeil, and Allen)

ods that this terminology has not been generally adopted, although, on the S-curve basis, it is logical.

The temperature ranges for  $Ar'$  and  $Ar''$  at various rates of cooling were shown in Fig. 7 for different C contents. By the thermal-analysis method used in getting these curves, there is no clear distinction between the transformation to all martensite or to bainite, and so both changes are lumped as  $Ar''$ . In a certain temperature range both changes are possible.

However, there is a very real differentiation between the two, as was brought out on p. 47.



What is almost invariably sought through hardening by quenching is martensite, as pure and free from bainite as possible, because the products obtained by tempering bainite fall short of the properties of those obtained by tempering martensite. For practical reasons, it is often necessary to put up with a martensite more or less contaminated with bainite, but that does not alter the fact that we generally prefer all martensite when we can get it.

Wilks, Cook, and Avery<sup>2</sup> emphasize the inferior "ductility, impact and other test values" resulting from quenching to bainite rather than martensite and tempering each. The toughness of untempered bainite (excellent) in a small range of relatively high hardnesses, to be discussed under austempering, is not so excellent when it is tempered and the product is compared with tempered martensite. The as-quenched strength and hardness of completely martensitic and of slack-quenched structures in C steels (50% martensite) are shown by Herres<sup>3</sup> in Fig. 145, and after tempering,<sup>1</sup> in Fig. 146. Increasing emphasis is put on the desirability of quenching to all-martensite.

**The Pressure Effect.** It is pointed out later (p. 314) that transformation to martensite is dependent not only on the inherent sluggishness of the austenite that has been retained, and the rate of cooling which provides the falling temperature necessary for completion of the transformation to martensite, but also on the *pressure* exerted upon the austenite. When a large section is quenched, the outside transforms to martensite, with a volume change. This alters the pressure on the underlying untransformed austenite. If martensite is formed at a different rate in one case than in another, so that the pressure on the untransformed austenite differs, that austenite will have different behavior as cooling proceeds.

The temperature gradients at chosen distances from the surface will not be alike in pieces of different section. Hence, to find the depth of hardening produced by a given quench of a given steel, conditions must be chosen so that the temperature gradient in the quench, and hence the pressure, exactly duplicates that of the section whose hardenability is sought. An alternative way of looking at this is that there must be plastic deformation of austenite by shear in order that martensite be produced.

This point of view is based on the observations that plastic deformation of unstable austenite in 18:8 induces the formation of martensite, but that application of hydrostatic pressure, without plastic deformation, does not. The proportional limit of austenite is rather

low, so that a tiny amount of plastic deformation could be rather readily produced. It is postulated that only tiny deformation is required.

Since the source of such deformation must be pressure, the two concepts are not far apart; nor is it material, from the point of view of practical utilization of the phenomena, which one ultimately survives.

The only sure way to establish identity of pressure or plastic deformation in a test such that the test will be reliable is to use the actual section. In small sections, a fair approach to identity can be secured by less time-consuming methods, such as the Jominy end-quench technique, but for heavy sections, say those in the neighborhood of 4 in. diameter, reliable results are not obtained.

**Appraisal of Hardenability.** Hence, for true appraisal of hardenability of large sections one must reproduce the degree of homogeni-

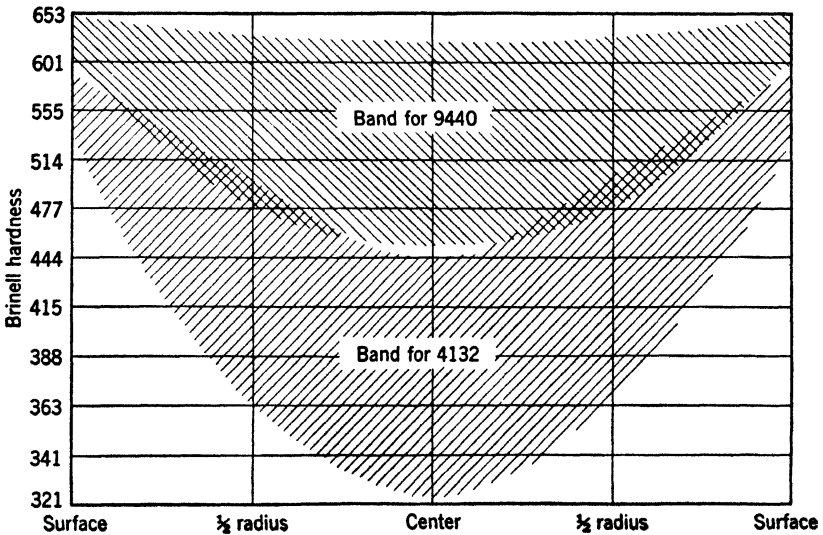


FIG. 147. Maximum and minimum hardness for 22 heats of old and alternate steels for steering knuckles. 1 3/4-in. rounds were normalized at 1650°F, water-quenched from 1550°F, sectioned, and their hardness surveyed.

zation of austenite and the quenching temperature and the intensity of quench to be used in the commercial hardening of the steel in question, and must use the actual large sections to be employed so as to get at the interior and evaluate the effectiveness of hardening in the interior.

This evaluation should not stop with a mere determination of the hardness gradient, but should be supplemented by microscopic or other examination to show how far the production of an all-martensite structure has extended into the interior. Tempering the quenched section under definite conditions and taking tensile or notched-bar impact specimens, at various depths, can be used as a criterion of slack quenching, since the strength, and especially the notched-bar behavior, serve as rather sensitive indices.

Neither the microscopic nor the notched bar nor other mechanical property evaluation is commonly made. Most of the available information rests on hardness alone. An example is shown in Fig. 147 from Clark et al.<sup>4</sup> The question at issue was whether N.E. 9440 in 1 $\frac{3}{4}$ -in. round would harden as effectively as 4132 which had previously been used. Figure 147 is a good one with which to start discussion of hardenability, for it brings out the scatter existing among different heats, all meeting chemical specifications. Since the center of the bar from the least hardenable heat of 9440 was as hardenable as that from the most hardenable heat of 4132, the 9440 appears to be a suitable substitute from the point of view of hardenability.

**Actual Determination.** Data on individual heats of several SAE steels obtained by actual quenching and sectioning are given by Klain and Lorig.<sup>5</sup> Bain<sup>6</sup> collects many such curves for a large variety of steels, using single heats. Figure 148 shows the effect of C. However, Fig. 149 from Bain for 1040 is not entirely consistent with Fig. 148; that is, individual heats behave differently. Figure 150 shows that 3140 hardens to the center in 1 in. diameter, but not in 2 in. diameter.

In Fig. 148, the surface-to-center hardness curves of 1035, 1045, and 1075, in the  $\frac{3}{4}$ -in. size, give a striking picture of the differences in hardenability. One half of the complete curve, say the left-hand half, will suffice. In deeply hardening alloy steels, the curves from a larger section would be needed, for a simple comparison. For precise comparison, some method of spreading out the curve and making it cover all degrees of hardening, from full martensitic down to no hardening at all, would be helpful.

Such a spreading of the curve is achieved, and the tedious technique of quenching and sectioning a range of sizes has been to some extent replaced by a scheme which gives much the same information on a single bar, and the information from which can be *approx-*

*mately* interpreted in terms of bars of different (not too large) diameters. The scheme also makes for precision in that the stretching out of the hardness-distance curve makes more easy a differentiation between steels of slightly varying hardenability.

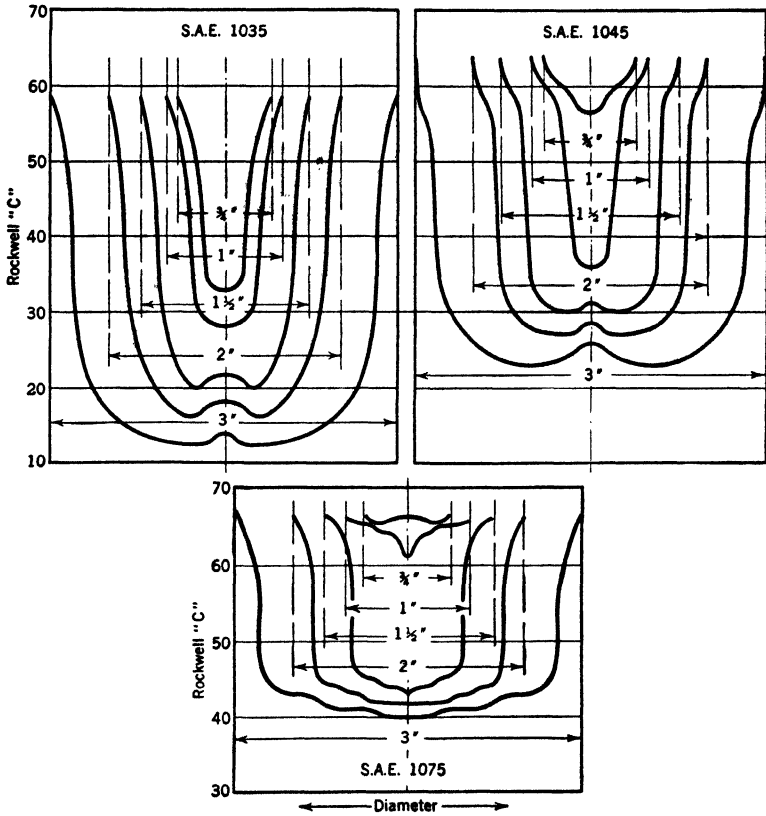


FIG. 148. Effect of diameter on depth of hardening in water-quenched fine-grained steels. (Burns, Moore and Archer<sup>7</sup>)

	Composition, %		
	SAE 1035	SAE 1045	SAE 1075
C	0.37	0.48	0.78
Mn	0.78	0.78	0.77
Si	0.28	0.27	0.19
P	0.016	0.016	0.028
S	0.082	0.030	0.031

**The Jominy End-Quench Test.** This scheme, due to Jominy and Boegehold, uses a 1-in.-diameter bar, 3 or 4 in. long, quenched at one end only by a strong stream of cold water applied for 3 min (Fig. 151). The cylindrical part of the bar is kept out of contact with

water. Thus the quenched end cools at a fast rate, like that at the surface of the smaller sections in Figs. 148–150, coming down to the  $M_s$  temperature in a fraction of a second, while the other end cools at a slow air-cooling rate. At intermediate distances the surface cools at intermediate rates.

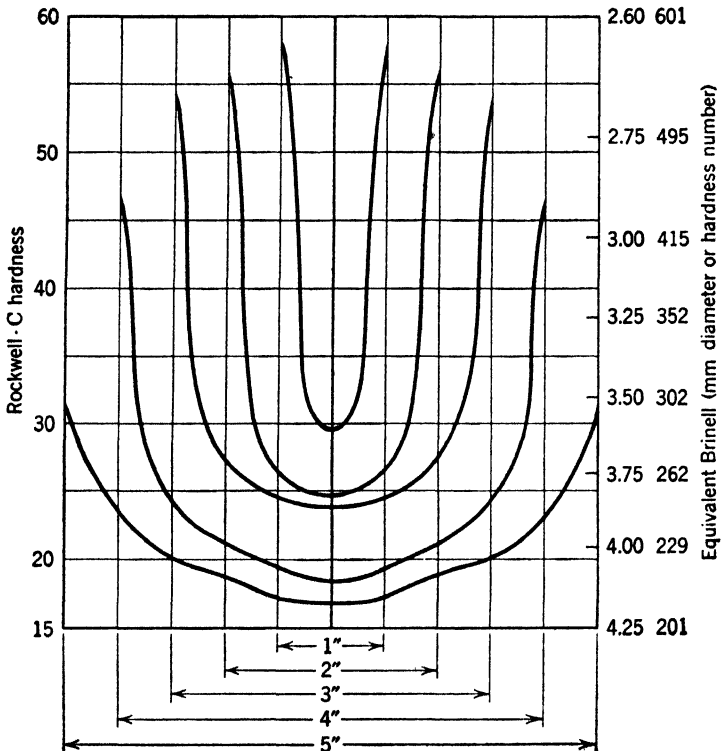


FIG. 149. Hardness distribution in quenched round bars of SAE 1040 steel. Vigorous water quench from  $1525^{\circ}\text{F}$ . (Courtesy W. M. Lindsey and E. L. Roff, South Chicago Works, Carnegie-Illinois Steel Corp.)

Some one of these surface rates will approximate the *center-cooling rate*, through the pearlite range, of any ordinary section. A small flat is ground on the cylinder length, a hardness traverse made, and the hardness–distance curve is plotted. Rockwell “C” is usually used for hardness, and the distance between impressions is usually  $\frac{1}{16}$  in. Metallographic examination and measurement of distances to 0.005 in. give a more refined evaluation. The hardness readings may be taken spirally, to allow better spacing.

The degree of homogenization, the grain size, and the quenching temperature in this test should be the same as it is planned to use in actual heat treatment. The structure of the steel before austenitizing should be stated, and the austenitizing time and temperature given.

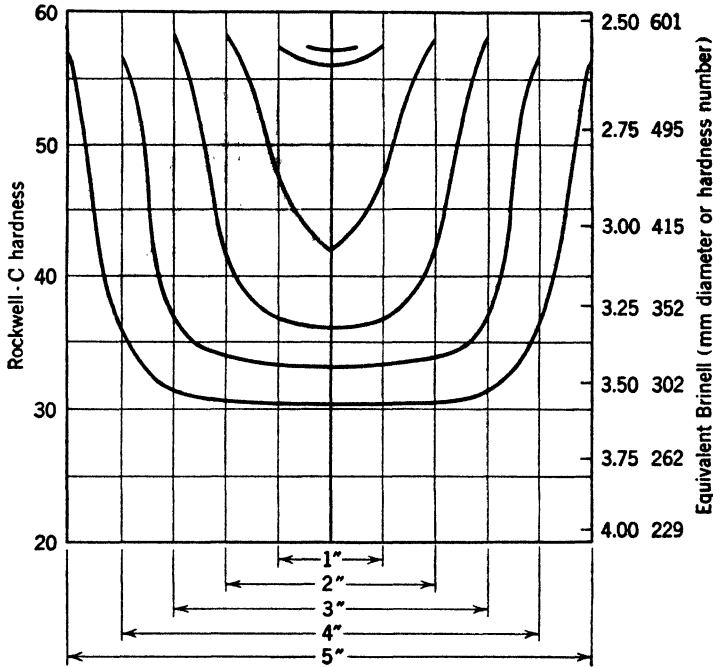
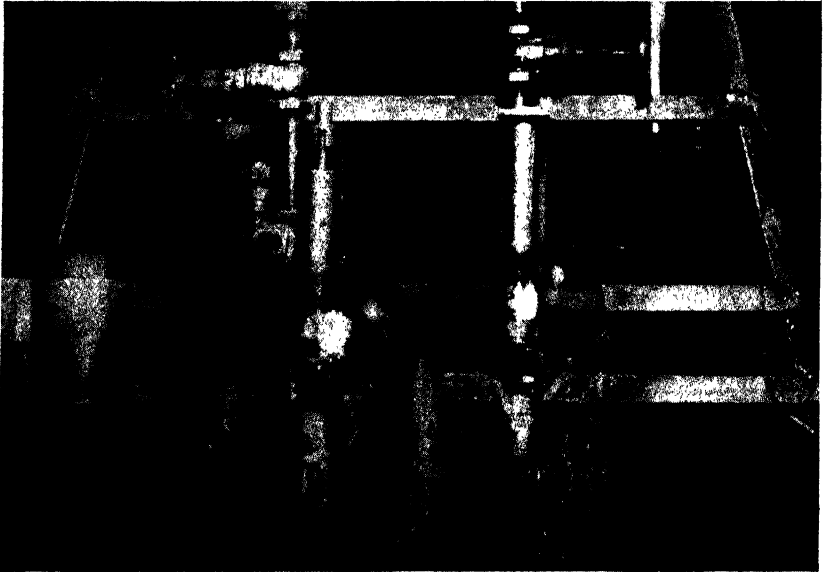


FIG. 150. Hardness distribution in quenched round bars of SAE 3140 steel. Average water quench from 1525°F. (Courtesy W. M. Lindsey and E. L. Roff South Chicago Works, Carnegie-Illinois Steel Corp.)

If we wish to establish that two heats of the *same* steel have similar hardenabilities, if the Jominy curves for the two lots, similarly homogenized and end-quenched at the same temperature, fall one upon the other, the proof is good, without necessity of recourse to metallographic examination. But, when different steels are being studied to see if one can be substituted for the other, we must recall that at the position on the bar where martensite and bainite have both been formed hardness values are not discriminatory, and additional data, obtained by metallographic examination, as to actual martensite percentage at the different distances are needed, because

bainite responds to tempering differently from martensite, and, when tempered, has poorer properties than tempered martensite.

**S-Curve Comparisons.** If the S curves for the two steels being compared were alike in the bainite region, and the temperatures for formation of martensite were the same, then the quenched struc-



(Battelle)

**FIG. 151.** Apparatus used for quenching end-quench hardenability bars. At the left is the standard 1-in.-diameter bar 3 in. long held by a stainless-steel threaded fixture; at the right is the standard 1-in.-diameter bar 4 in. long with a machined shoulder for supporting the bar in the quenching apparatus. Both bars are quenched on the end by a stream of water discharged through a  $\frac{1}{2}$ -in.-diameter orifice  $\frac{1}{2}$  in. from the end of the bar. The free height of the quenching stream is  $2\frac{1}{2}$  in.

tures of the two steels would also be alike. S-curve data are especially valuable for just this purpose, but the bainite portion of the S curves is the one ordinarily least completely evaluated. Figure 152 brings together the S curves (not corrected for continuous cooling) and the Jominy curves for 3140 and 9640. The similarity in the time delay at the pearlite nose, and the fair similarity of the bainite bays lead to the prediction that the two should respond quite similarly to similar cooling rates. The Jominy curves verify this expectation.

Hence, actual hardenability determinations are usually made by the end-quench technique. The effect of alloy elements is shown by S curves and by Jominy data. The S curve shows the general effect for a given type of steel. The Jominy curves are more easily deter-

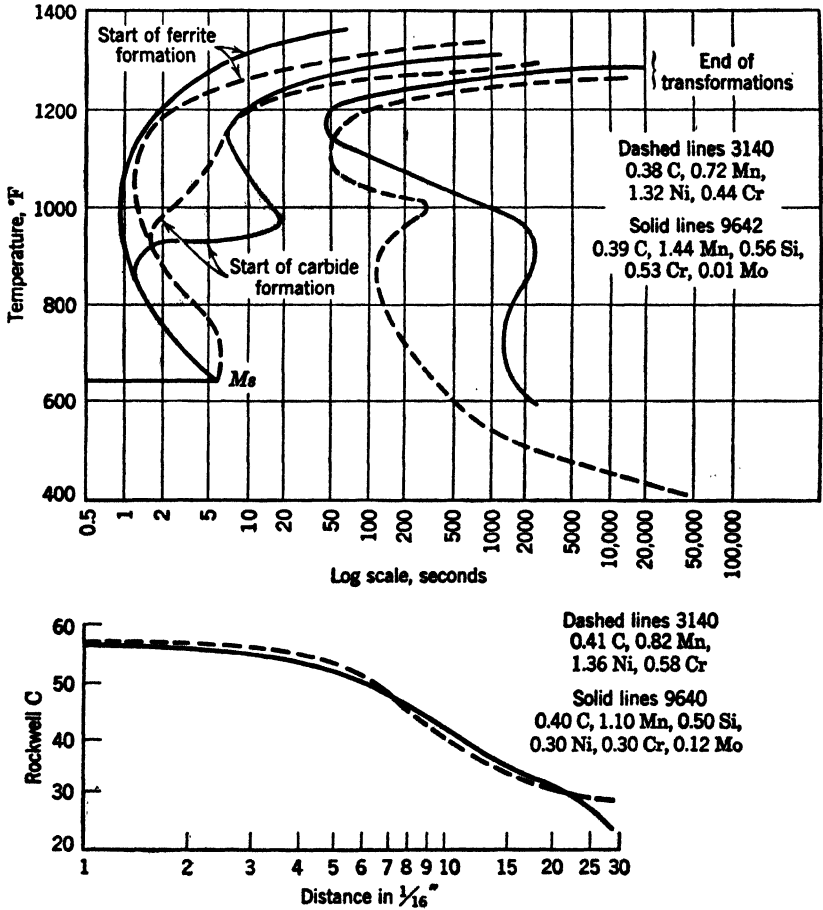


Fig. 152. S curves and Jominy curves compared.

mined, and hence more convenient for showing the differences resulting from known changes in composition and from those unknown factors that make the hardenability of two heats of like composition and treatment often vary widely.

Publications giving Jominy curves for individual steels are legion. Very often such curves show wide disagreement among steels usually classed under a single AISI, SAE, or N.E. designation.



**Hardenability Bands.** Much interest is therefore taken in developing the "hardenability bands" for different classes of steel. Ekholm<sup>8</sup> has summarized the spread met, within the normal chemical limits of composition, in the Jominy curves of some of the 87 currently used steels, for which hardenability bands have so far been

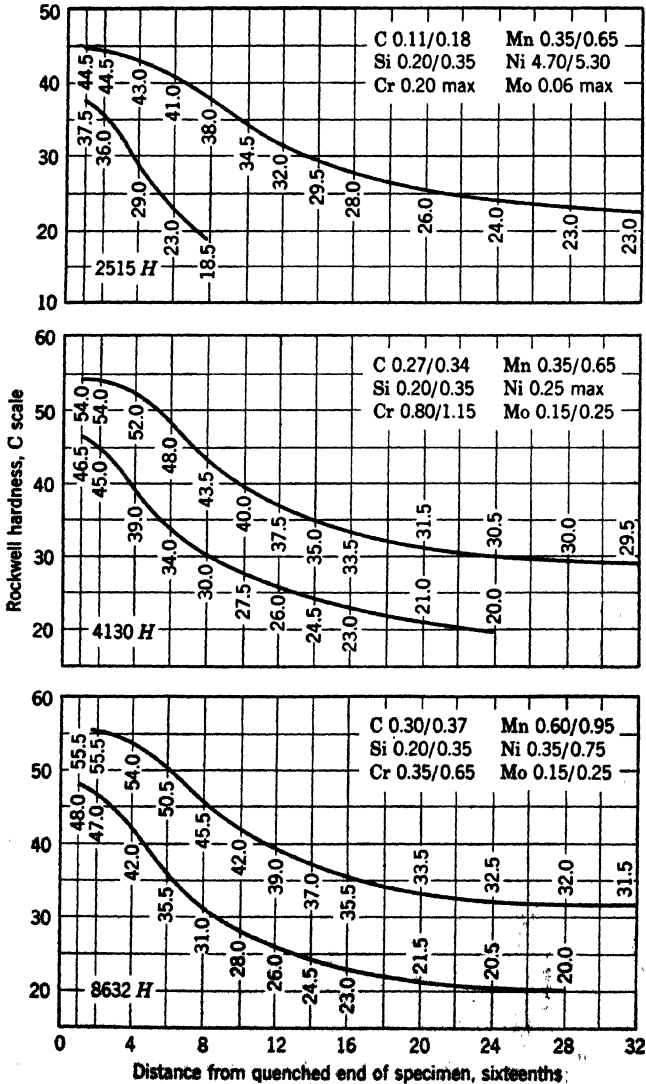


FIG. 153. Jominy bands for compositions shown.

determined by the American Iron and Steel Institute.<sup>86</sup> Some of the bands are shown in Figs. 153-154.

The separate curves for individual heats of three steels,<sup>9</sup> Figs. 155-156 and Fig. 157, from Caine<sup>10</sup> show a spread in cast steels, also show

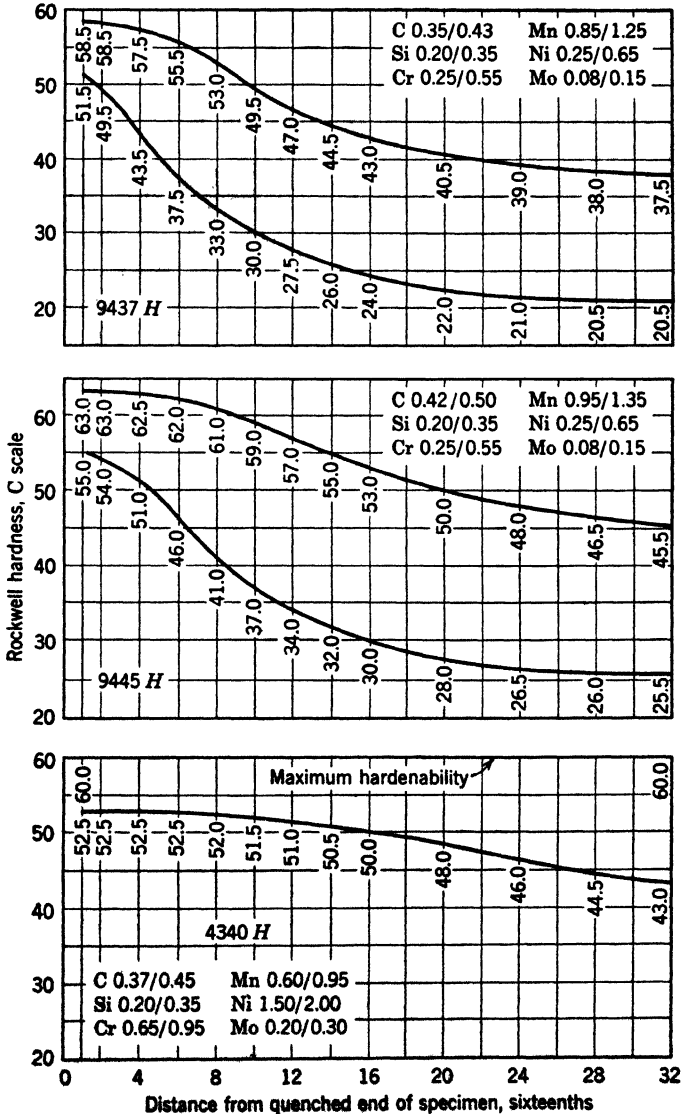


FIG. 154. Jominy bands for compositions shown.

that an "average" curve for wrought 2320 falls within the band for cast 2320. The notations at the top of Fig. 157 are referred to later.

Figure 158 shows the scatter<sup>11</sup> for a few lots of each of two steels, plotted on a frequency basis.

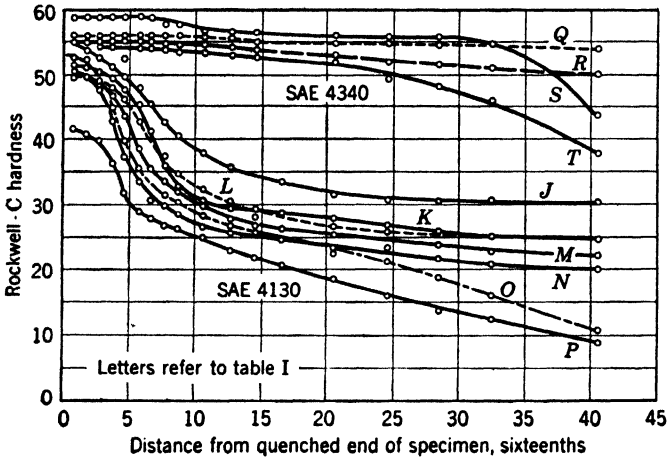


FIG. 155. Hardenability of typical heats of 4130 and 4340 steels.

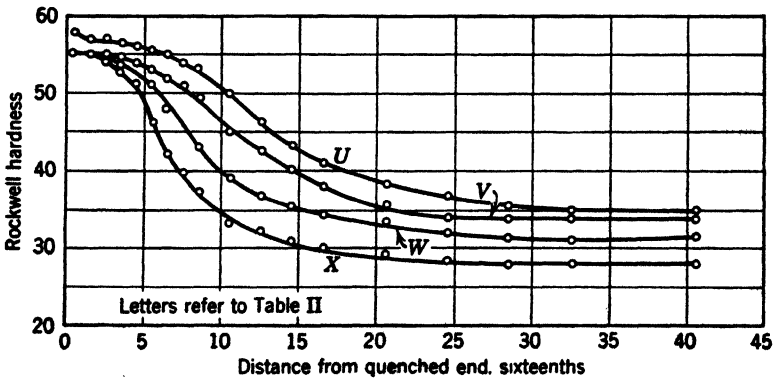


FIG. 156. Hardenability of typical heats of 4140 steel.

As Fig. 148 indicated, small differences in C content greatly affect hardenability of steel. Figure 159<sup>12</sup> shows this for the 4600 series. Note that the hypereutectoid (1.03% C) steel appears less hardenable than the eutectoid steel. The reason is that, with the treatment given, the C was not fully dissolved. Figure 160 shows,<sup>13</sup> for a particular heat of 5100, the important effect of austenitizing

temperature in increasing solution of C, and thus increasing hardenability.

**Segregation.** If the steel is inhomogeneous, either through general banding or chemical segregation, as may occur at the center of an ingot, anomalous hardenability will result. Parke and Herzig<sup>14</sup> show cases in 1330 and 4140 where banded steels gave deeper hard-

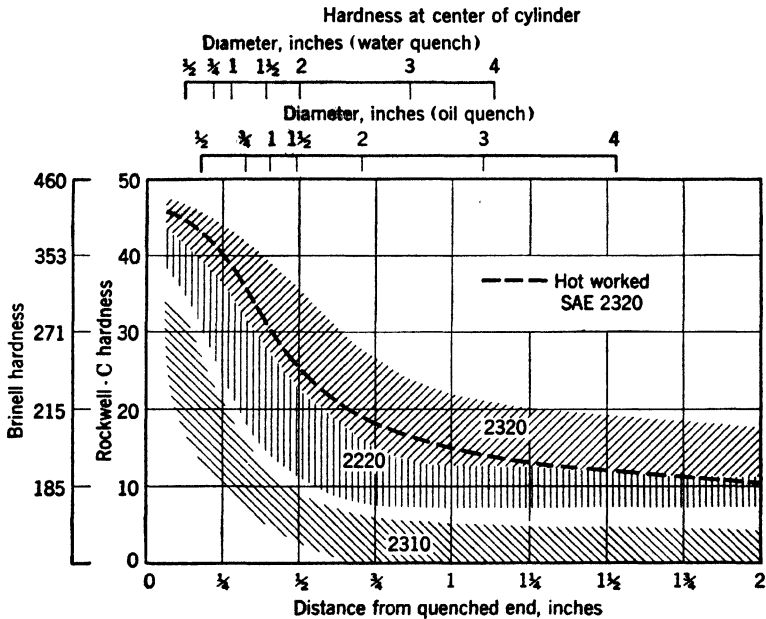


Fig. 157. Hardenability bands of nickel-carburizing cast steels.

enability in the Jominy test than they did after the homogenization was increased. The explanation is that the C and alloy content of some of the bands was so much higher in the unhomogenized condition that the effect overbalanced that of the locations poorer in C and alloy. Metallographic examination clarified the anomaly.

Walker and coworkers<sup>15</sup> found banding and center segregation in 4150, in this case resulting in low hardenability at the center. A Jominy bar, cut from a 2 1/4-in. round so that one side of the Jominy bar was at the center of the round and the other close to the surface, gave quite different hardness-distance curves for the two sides. This emphasizes that, like any other test, sampling comes into the Jominy test, and, in case of center segregation, a Jominy bar turned down from a larger bar will tell the hardenability at the location in the

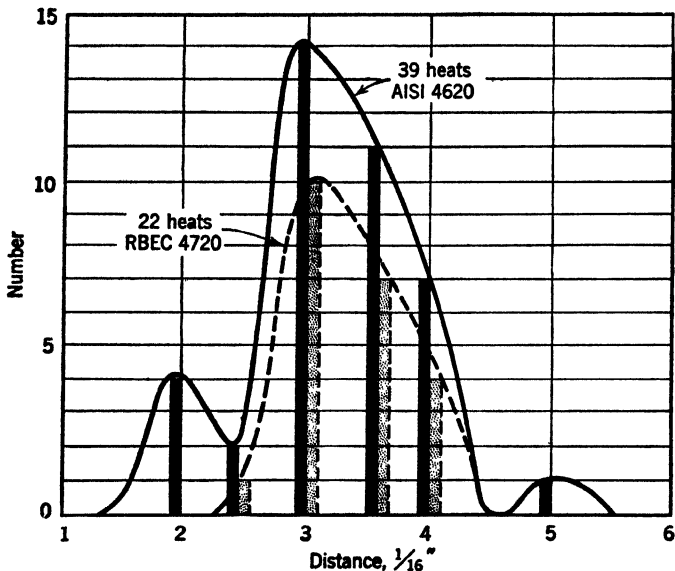


FIG. 158. Frequency distribution of hardenability (distance on end-quenched specimen to RC-30) of several heats of 4620 and 4720. Samples normalized at 1650 and end-quenched from 1650°F.

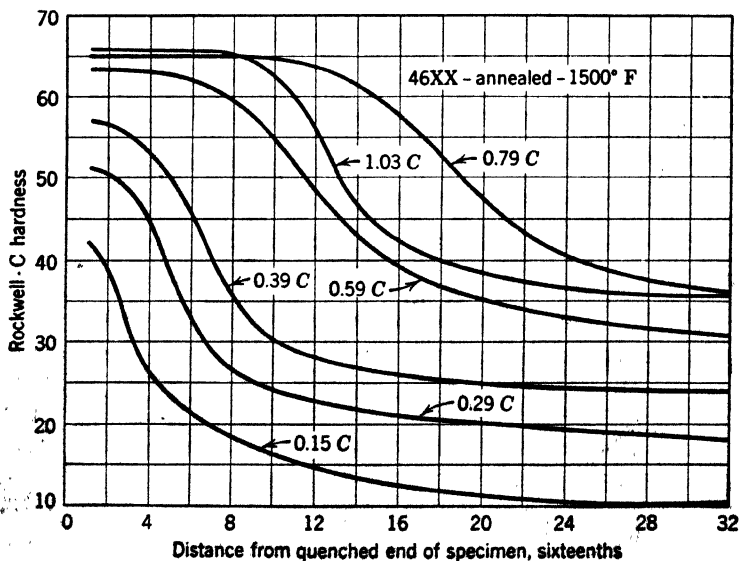


FIG. 159. Effect of carbon content on the hardenability of the 4600 series from the annealed condition.

original bar along which the hardness readings are taken, but at no other.

When the user is satisfied that a certain quenched hardness, whatever its structure, is satisfactory, or when he knows that the steels being considered have like S curves, so that a given hardness does connote a definite structure, he may describe hardenability by the

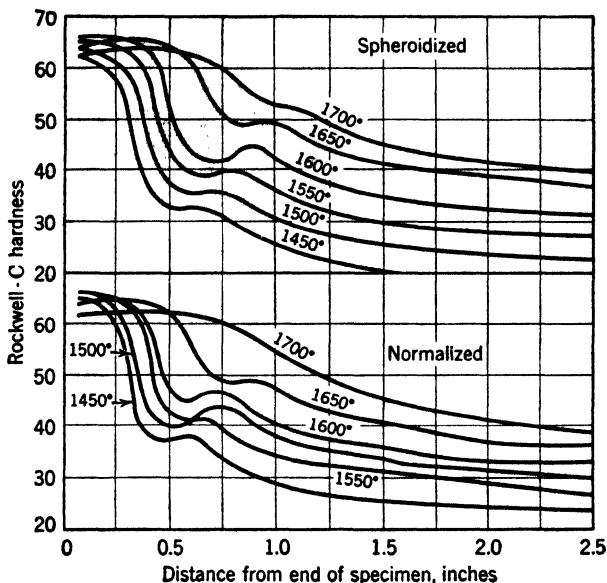


FIG. 160. Effect of quenching temperature on hardenability of 5100 steel after 40 min at temperature.

distance along the Jominy bar at which this selected hardness obtains. Thus "J 50-24" is code, or shorthand, for saying that on the Jominy curve, a hardness of 50 Rockwell C is shown at  $2\frac{4}{16}$  in. from the quenched end. Compare Parker.<sup>41</sup>

The distances on the Jominy bar, end-quenched with high-pressure water, relate to rather definite cooling rates. These cooling rates may be compared with the "critical" cooling rates which just avoid production of ferrite, pearlite or cementite, or bainite, and retain 100% austenite. However, as Fig. 161 shows, different experimenters differ widely as to the actual critical cooling rate for plain C steels, though the contours of the curves are similar.

Several investigators<sup>16, 17</sup> have studied the cooling rates at different positions along the Jominy bar. Hill<sup>17</sup> shows a spread or un-

certainty in the data, at  $\frac{1}{8}$  in. from the quenched end, ranging from under  $200^\circ$  per second to over  $300^\circ$  per second, and considerable disagreement down to  $\frac{1}{2}$  in. from the quenched end. Kopecki<sup>37</sup> shows even greater disagreement (and describes a wedge-shaped specimen for shallow-hardening steels which has obvious advantages). Translocation of Jominy distances into fast-cooling rates is therefore attended with considerable uncertainty.

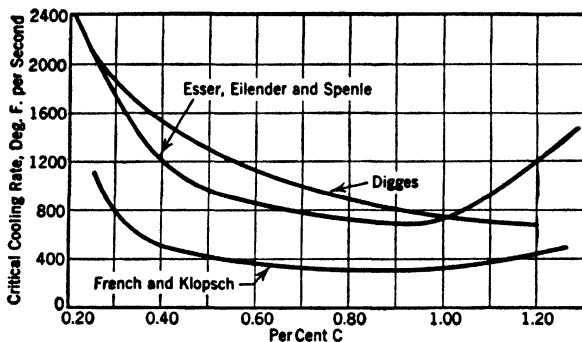


FIG. 161. Variation of critical cooling rate with C content, as determined by different investigators.

**Jominy Distances versus Cooling Rates.** As is shown at the top of Fig. 157, a given distance on the Jominy bar is supposed to denote the diameter of a bar, quenched all over, in still water or still oil, which will harden (to 50% martensite) to the center. The early charts for plotting Jominy curves, put out by the ASTM, carried a scale giving what were then assumed to be the cooling rates at different distances along the Jominy bar. Thus the rate at  $\frac{1}{16}$  in. from the quenched end was given as  $600^\circ$  per second; at  $\frac{1}{2}$  in.,  $30^\circ$  per second, etc.; down to  $4^\circ$  per second at  $2\frac{1}{2}$  in. (see Fig. 167). The later investigators showed that those cooling rates were not accurate. (Neither may the "hardenable diameter" scalings be taken at full value, for they were based on the early fallacy that the proper criterion of hardening was a structure containing only 50% martensite.)

More recently, the cooling time in seconds required to cool from  $1300^\circ$  to  $900^\circ$  has been taken, instead of the cooling rate at  $1300^\circ$ , for mathematical calculations on the effect of cooling on depth of hardening. Still another criterion is the "half-temperature cooling time."<sup>38</sup> As is discussed later (Chapter 13), the mathematical cal-

culations have led to few if any useful results in relation to hardenability.

Hence, in the present-day plotting of Jominy curves, the water and oil size relationships and the cooling rates are not often shown. The things plotted are those actually observed, the hardness and the distance.

In a few highly commendable studies, observations of the microstructure are made, and percentage of martensite is plotted against distance. Such plots are by far the more informative.

An uncertainty in the translation of Jominy distance to the size in which a given steel will harden comes in the fact that, for reproducibility of results, the end of Jominy bar is pressure-quenched with water, a much more severe quench than is ordinarily in actual use on commercial parts. Varying intensity of the quench, with consequent variation in cooling rates, alters the depth hardened.<sup>35</sup> From a practical point of view, one finds by experiment the size that hardens to the depth desired in the quench he will actually use, on a given lot of steel. A Jominy curve on that steel serves as a base line for comparison of other lots and other steels. If the hardness and the microstructures of the base-line steel are identical at equal Jominy distances, the response to the actual quench will be identical. Hence the fact that the Jominy quench is an unusually intense one does not detract from the value of the test.

**Intensity of Quench.** However, in the attempt to develop concepts and nomenclature that will generalize the discussion of hardenability, the intensity of quench has been taken into account. In order not to blame the steel for the failure to apply, in practice, as severe a quench as the quenched end of the Jominy bar gets, comparison of steels is made on the basis of how they would behave if the quench, applied to different-sized pieces of the steels, were the "ideal" one, that is, the most intense quench possible. In this hypothetical quench the diameter that would harden through, likewise hypothetical, is termed the "ideal diameter." Again, we have to define what we mean by "hardening through," that is, whether we want all-martensite or will be content with 50% martensite, before the term "ideal diameter" has meaning. To be logical, when we are demanding the most powerful quench, we should also demand complete response to the quench, that is, 100% martensite, but the early investigators did not see fit to demand complete hardening and used the 50% criterion. Thus the "ideal diameter," as ordinarily discussed, is much greater than the diameter that will harden in an



actual quench to the desired product, martensite. It is a fictitious figure.

A picture of the relative intensity of brine, water, and oil quenches is obtained in Fig. 162, from Clark and Richards<sup>18</sup> for the behavior of a particular steel. Particularly striking is the hardening (to 50%

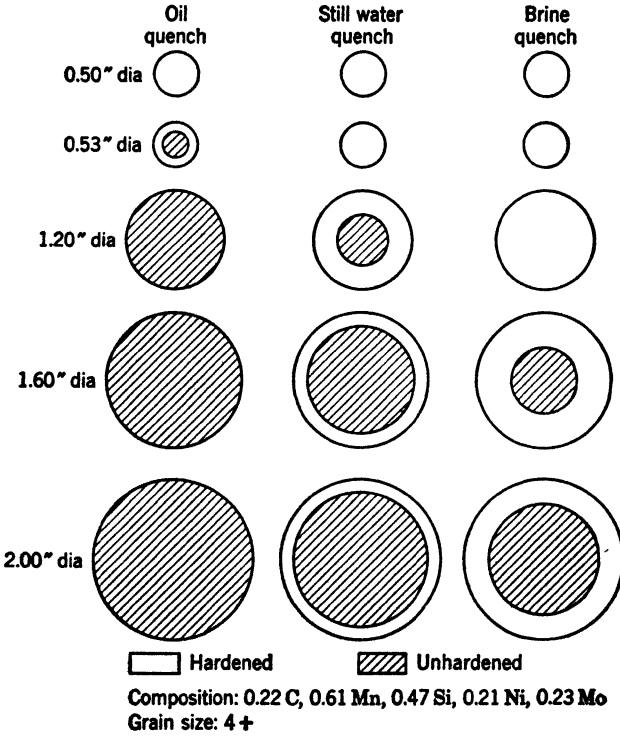


Fig. 162. Depth hardening, to 50% martensite, as affected by the quenching medium.

martensite) in oil of a rod, 0.50 in. diameter, against the retention of a considerable unhardened core when the diameter is increased only to 0.53 in.

**“Ideal Diameter.”** The symbol used for “ideal diameter” is  $D_I$ ; that for intensity of quench is  $H$ . An  $H$  of 10 is supposed to be the most vigorous one could possibly obtain. Brine solution under strong pressure or violent agitation, quenching a scale-free steel, is supposed to show an  $H$  of 5. Water quenches under pressure, as at the end of the Jominy bar, around 2; still water, around 1; and oil quenches from around 0.25 to, say, 0.75, depending on the agitation. Accord-

ing to Grossmann<sup>19</sup> the observed hardened diameter (to 50% martensite) in a quench of given intensity is related to  $D_I$  as shown by the curves in Fig. 163. Note how much greater is  $D_I$  than the diameter hardened in normal quenches. It is not easy to evaluate  $H$  accurately for a commercial quench.

Hodge and Orehoski<sup>20</sup> gave Fig. 164 for  $D_I$  (note that the most vigorous quench possible is involved in  $D_I$ ) when the all-martensite

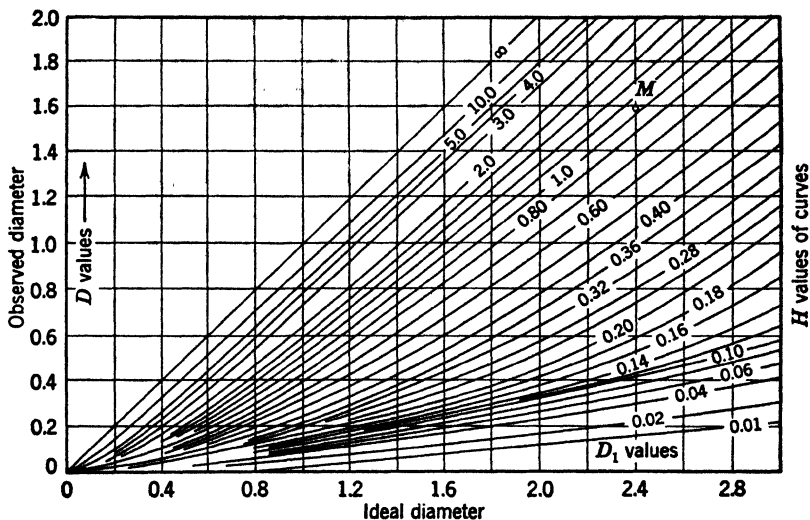


FIG. 163. Relation between actual observed hardenable diameter and "ideal diameter," postulated to be obtainable in the fastest possible quench, in terms of "intensity of quench,  $H$ ."

criterion or those of other percentages of martensite are used instead of the older 50% criterion.

The fictitious value  $D_I$  is of theoretical rather than practical interest. It was used when an attempt was being made to calculate hardenability from chemical composition. When the  $D_I$  figure is wanted, it is ordinarily taken from the curve, Fig. 165, as revised by Hodge and Orehoski<sup>21</sup> in 1946, the actual observations being made on Jominy bars.

The conversion curve of Fig. 165 was developed from tests on a series of low-alloy steels of the N.E. type. Grange and Garvey<sup>22</sup> point out that the ratio of  $D_I$  for all-martensite to 50% martensite varies between low-alloy and plain C steels, and Hollomon and Jaffe<sup>23</sup> point out that, when some other criterion than all-martensite

is used,  $D_I$  will vary according to what other constituents accompany martensite. The balance other than martensite may be ferrite and

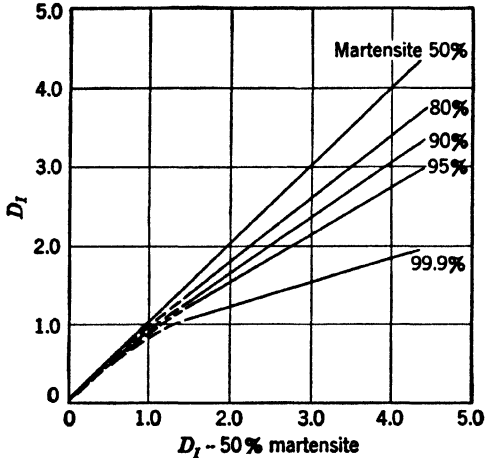


FIG. 164. Average relationship between hardenability based on higher percentages of martensite and 50% martensite.

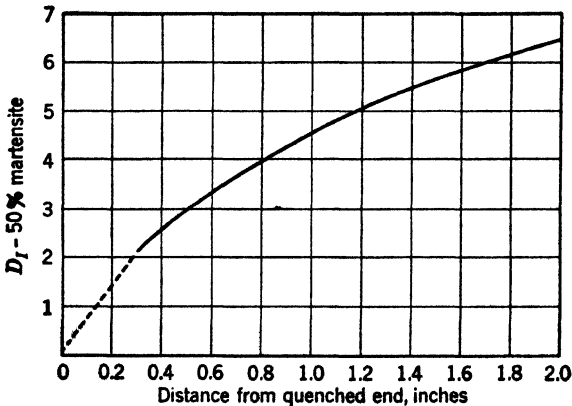


FIG. 165. Curve for conversion of Jominy distance to "ideal diameter."

pearlite, or bainite, according to the steel and the cooling rate down to  $M_s$  and through the  $M_s$ - $M_f$  range.

Because of these uncertainties the practical primary criterion for comparison of depth hardenability of different heats or different steels is Jominy distance, not  $D_I$ .

**Hardness of Martensite versus Carbon Content.** The hardness of an all-martensite structure depends on C content. Hodge and Orehoski<sup>20</sup> give Fig. 166 for the relation between the two. The upper curve, for all-martensite, appears to hold for most of the conventional steels used for heat treating. The lower curve shows wider scatter, and the relation may well vary beyond the limits shown, in steels other than the low-alloy series studied by Hodge

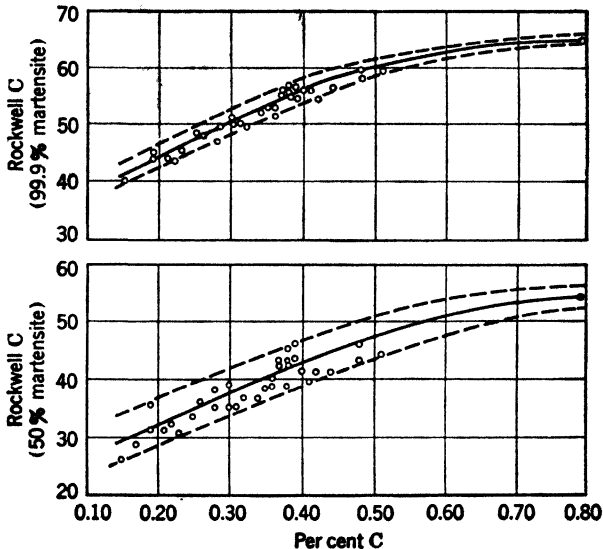


FIG. 166. Hardness of martensite products as function of C content.

and Orehoski, since a steel which quenches to a mixture of 50% martensite and 50% hard lower bainite will be harder than one quenching to 50% martensite and 50% of the softer ferrite and cementite. Hence mere hardness, even of steels with the same C content, is not a satisfactory criterion when steels of different types are being compared. Hardness alone may serve when the problem is the comparison of steels with very similar S curves, or of different heats of nominally the same steel. Otherwise, the hardness values need to be supplemented by metallographic study of the structure.

When (and only when) experience has shown that the core of a quenched part does not need to be fully hardened, and can tolerate the presence of ferrite and pearlite, or of bainite, as the case may be, limits may be set as to hardness, and it may then be demanded that an acceptable steel show a particular hardness at a given Jominy

distance. Focke,<sup>24</sup> in Fig. 167, has set such limits for particular cases. The limits seem pretty wide; the so-called "critical hardness" might be far too low for other applications.

This is brought out in Fig. 168 collected by Shepherd<sup>25</sup> from American Iron and Steel Institute data on some N.E. steels. In these tests oil-quenched rounds up to 5 in. diameter were tempered

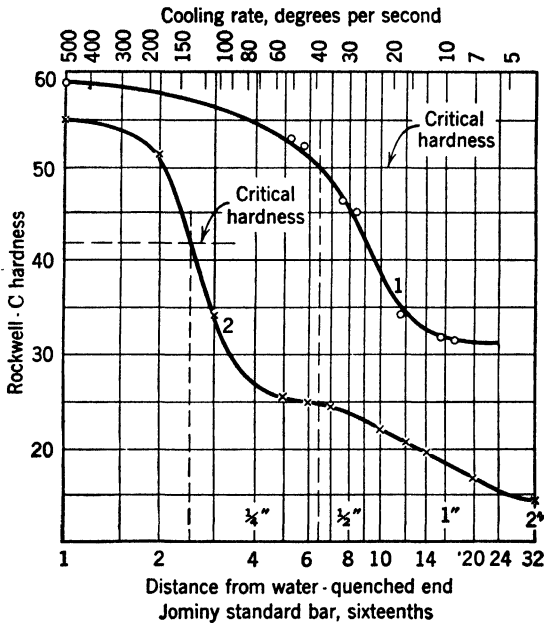


Fig. 167. 3140 and 1045 quenched from 1500°F as indicated. (Focke)

at 1000°, and 0.505-in. test specimens were cut from the midway position (in sizes above 1 in.). The yield and tensile strengths, so determined, are plotted. The "critical-bar diameter," that is, the diameter of bar that would harden on oil quenching to "50% martensite," which was assumed to correspond to the quenched hardness of 41½ and 45½ R.C., respectively, for 8630 and 8739, indicated for these steels by the Jominy test, is shown by arrows. The plots show that even at the midway position, to say nothing of what would have shown at the center, the "50% martensite" criterion includes a slack-quenched core which, after tempering, is of distinctly lower strength than a fully quenched, then tempered, structure.

The spread among individual heats is in evidence here. The Jominy tests on 8630 indicated a spread in "critical-bar diameters"

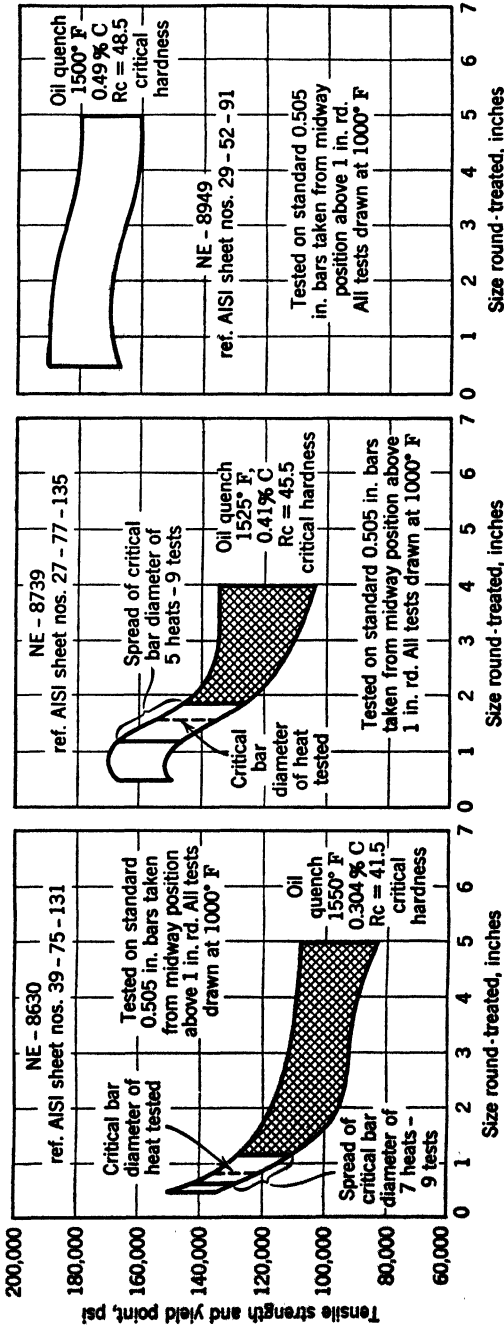


Fig. 168. Slack quenching of large sections of shallow-hardening steels.

of 0.7 to 1.1 in., and on 8739 of 1.15 to 1.8 in. This variability is reflected in the actual tensile results. The cross-hatched portions of the plots indicate such a degree of slack quenching as to eliminate such sizes for use where a center effectively strengthened by heat treatment is wanted.

Many such examples point to the inadequacy of the 50% martensite criterion in alloy steels. That criterion used in the early investigations was adequate for C and other shallow-hardening steels, but is inadequate and incorrect for the great bulk of the deeper-hardening steels that must be used when we seek to harden large sections. Bain<sup>88</sup> long ago realized and stated this. Crafts<sup>89</sup> puts it as follows: "Any correlation of hardenability with austenite transformation must be based on 100% martensite in order to simplify the problem and to evaluate hardenability in terms of real physical significance."

In shallow-hardening steels, a structure containing 50% martensite has the other 50% chiefly made up of ferrite plus pearlite. In such steels, when a quenched cross section is acid-etched, there is a visible difference in the color of the exterior hardened zone and the inner unhardened zone, so that measurement of the hard zone can be made without resorting to hardness measurements or microscopic examination of the structure (compare Fig. 162). If hardness measurements are made, it is found that the hardness-distance (periphery to center) curve drops very sharply at the boundary, and under the microscope the distance between an all-martensite structure and the 50% martensite structure is very small.

But as we add alloys and produce a more deeply hardening steel, the boundary shown on etching is no longer sharp and, as the next easiest determination, the hardness-distance curve is determined. In such a case, where slack quenching produces low-temperature bainite, which is hard, though not quite so hard as martensite, the hardness-distance curve is not so steep. Figures 149 and 150 showed hardness distance curves on 1040 and 3140, both water-quenched in various sizes. The sharp drop in the carbon steel and the more gradual drop in the 3140 are evident. The same thing, of course, occurs in the Jominy curves.

There are other factors, beside intensity of quench, that affect the hardenability and the Jominy curves. As Figs. 159 and 160 indicate, the structure of the steel prior to austenitizing for hardening, or for Jominy testing, affects the rate of homogenization of austenite, and so the behavior on quenching of austenite brought to

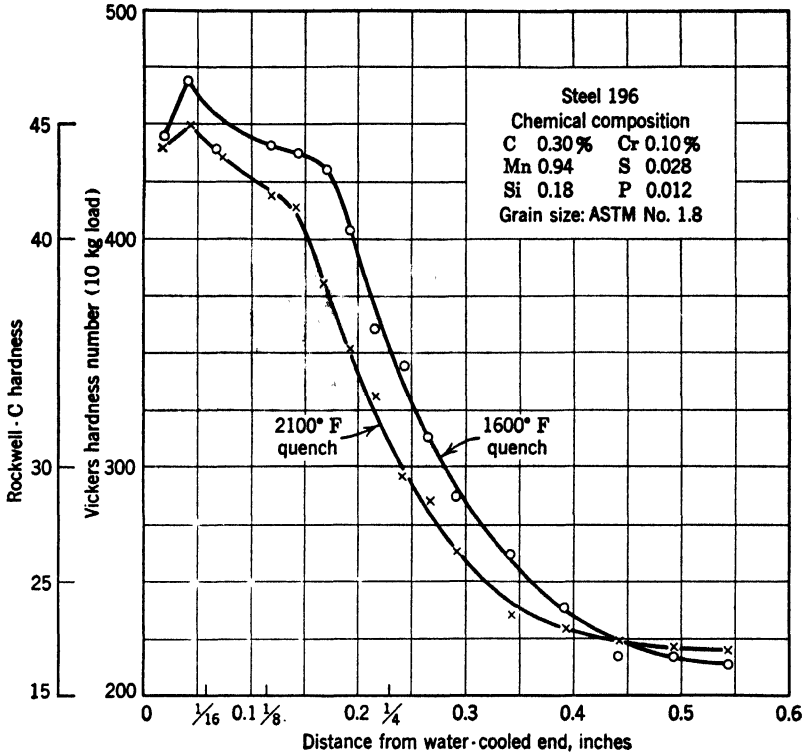


FIG. 169. Effect of quenching temperature on Jominy curves.

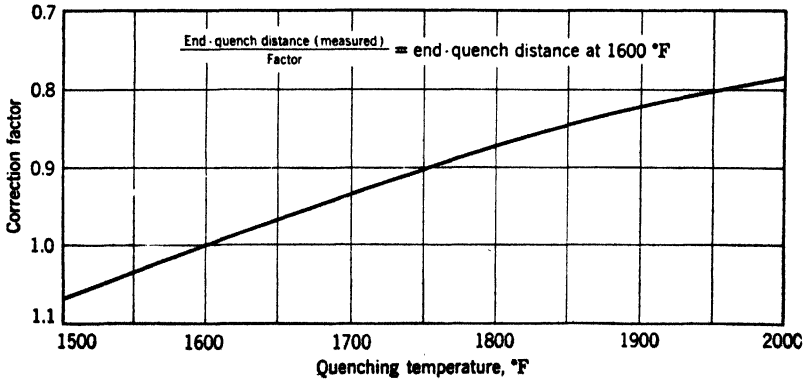


FIG. 170. Curve for correction of distance along end-quench hardenability bar for quenching temperature, 50% martensite criterion.



different degrees of homogeneity by heating at different temperatures and for different times depends on those factors.

For a particular steel Fig. 169 shows the observed differences<sup>28</sup> in the Jominy curves when the bar is end-quenched from different

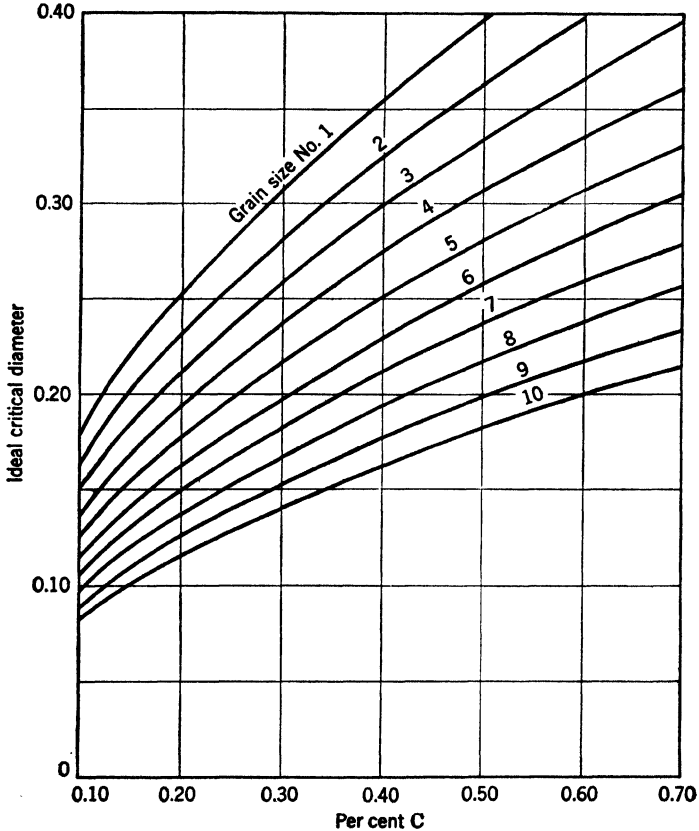


FIG. 171a. Ideal critical diameters for various C contents and grain sizes.

temperatures, and Fig. 170 shows a "correction curve"<sup>18</sup> for Jominy distances. This would not be expected to hold accurately for all steels, since much variation is shown in rates of homogenization.

Steels that are susceptible to grain coarsening at the temperature and time chosen for austenitization show quite different hardenabilities according to the austenite grain size existing when they are quenched. Grossmann<sup>19</sup> and Clark and Richards<sup>18</sup> (using  $D_I$  for 50% martensite as the basis of comparison) gave curves for this hardenability "correction factor," shown in Fig. 171a.

Erb,<sup>33</sup> however, finds that in hypereutectoid tool steels, as the C increases, increasing fineness of grain, as read by comparison with fracture grain-size standards, tends to reduce hardenability. If the carbides were taken into full solution, this would not be the case; hence Erb's curves, Fig. 171b, show the behavior at low austenitizing temperatures, which fail to produce grain growth and fail to take all the carbide into solution.

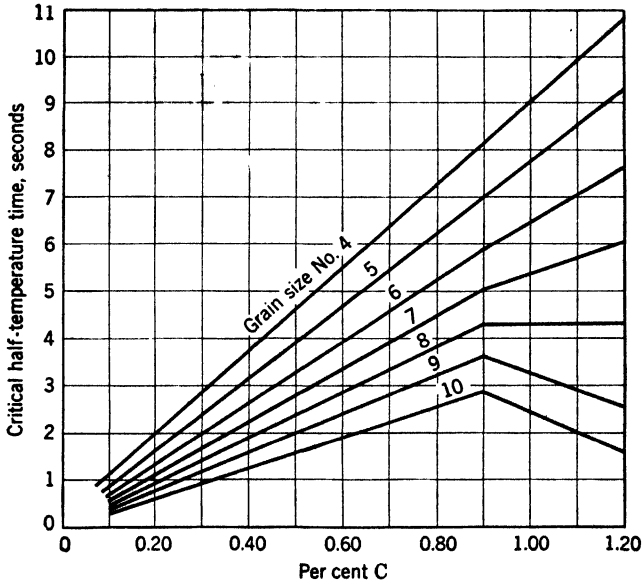


FIG. 171b. Relation of C content and grain size to hardenability. Half-temperature time computed from Grossman's ideal critical diameter by means of the equation  $D_I^2 = Kt$ , where  $D_I$  is the ideal critical diameter,  $t$  = half-temperature time, and  $K$  = constant = 0.2.

Kramer, Siegel and Brooks<sup>34</sup> find quite different correction factors for Si-killed steels than for Al-killed steels. Hence Fig. 171 is not universally applicable.

**Modified Jominy Bars.** Various modifications of the straight Jominy test may be applied for specific purposes. If the stock is smaller than 1 in. diameter, or is thin flat stock, a composite bar may be made up, using some other steel with a socket cut in it into which the piece to be tested is tightly inserted, making a bar of standard dimensions. Better to differentiate among shallow hardening steels in which the flat portion of the curve is short and the  $\frac{1}{16}$ -in. spacing not close enough to show just where the curve starts to drop, the bar

may be modified to the so-called L form (shown in Fig. 172) in which the quenched end, instead of being flat, has a conical hole, so

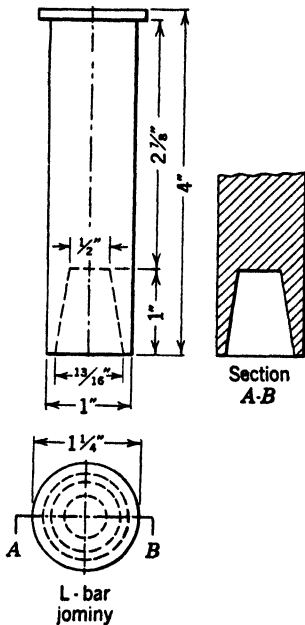


FIG. 172. Jominy specimen for very shallow-hardening steels.

that the hardened portion of the curve is longer; that is, the temperature gradient is made less steep near the quenched end. Regular and L bar curves for a shallow-hardening steel are shown in Fig. 173, from Rickett and coworkers.<sup>27</sup> Note the plotting of martensite percentage in the lower graph. When all Jominy curves are so plotted, much more useful information will be at hand than the literature now contains.

**Specimens for Shallow-Hardening Steels.** An alternative specimen for shallow-hardening steels is a 5-in.-long cone,  $\frac{1}{4}$  in. in diameter at the small end,  $1\frac{1}{4}$  in. at the other, pressure-quenched in brine all over, split, ground, and traversed for hardness, as described by Post and coworkers.<sup>28</sup> The severity of quench can be figured from such specimens more accurately than by the Jominy bar.

McMullan<sup>29</sup> avoids splitting the cone by cutting a wedge from a round or rectangular bar and quenching one face (Fig. 174). Hardness measurements are made on the hypotenuse. Kopecki<sup>37</sup> describes a wedge-ended bar, quenched on both faces of the wedge. Frericks and Rowland<sup>30</sup> made an exhaustive study of various bars and techniques for evaluating hardenability of shallow-hardening steels, including the carburizing and grinding down of Jominy bars to determine the behavior in the transition zone. They emphasize that the individuality of different heats is noteworthy in shallow-hardening carburizing steels.

In discussion, Dalrymple remarked that none of the conventional hardenability tests accurately predicted the core hardness of a carburized part in  $\frac{5}{16}$ -in.-thick pieces.

In some cases it is desired to match the temperature gradient of the Jominy bar more closely to that of the section to be quenched. Wilks, Cook and Avery<sup>2</sup> describe a double-end-quenched specimen

designed to give more readily applied data for such cases as the hardening of armor plate, which is quenched from both sides.

Their curve, Fig. 175, brings out another important point in the use of Jominy curves. A certain position, call it *A*, in the standard Jominy bar has quite the same cooling rate down to 1000° as does the center of a certain plate, call it position *B*, when quenched from

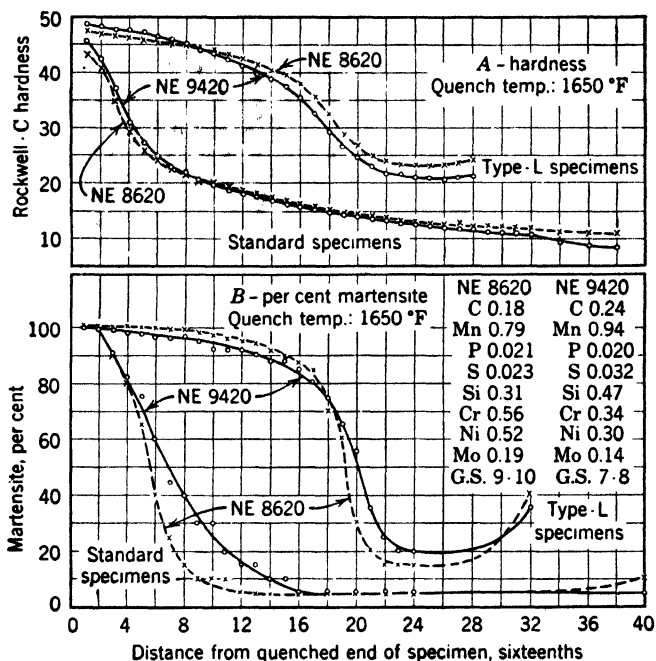


FIG. 173. End-quench hardenability of 8620 and 9420.

both sides in water, but the cooling curves, from 1000° down, deviate considerably. Hence the tendencies to form bainite rather than martensite can differ considerably on steels whose bainite S curves are cut differently by the cooling curves.

It takes rather large amounts of lower bainite in martensite to produce hardness much below that of pure martensite; hence, if we assumed that closely agreeing Jominy curves meant the same structures, we would often be wide of the mark. The assumption that equal hardness at Jominy position *A* and plate position *B* means that the same cooling curves have existed is not justified. Conversely one cannot take a Jominy curve and work it backwards to predict the structure at a given position in a piece actually being quenched in the

regular fashion. If the special Jominy bar is worked out to fit the case, with its *complete* cooling curve at some point the same as the complete cooling curve of the center of the actual piece, comparison

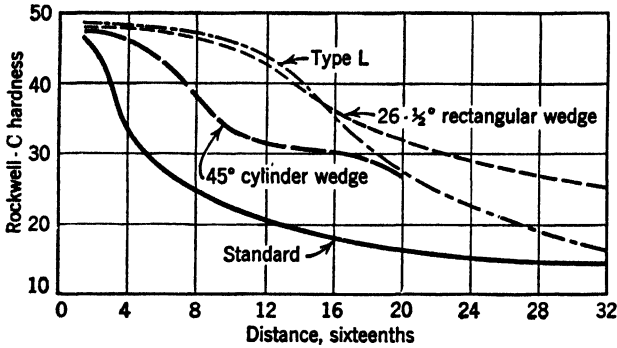


FIG. 174. Hardness curves of 8620 steel, No. 15, obtained from four types of test specimens.

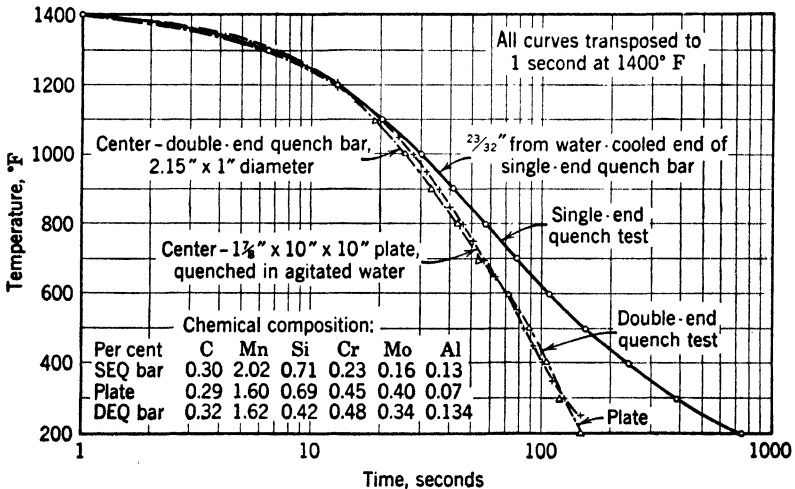


FIG. 175. Comparison of plate with single and double end-quench cooling curves. Quenched from 1650°F.

is valid for that particular position, if a "size effect," that is, a pressure effect, does not intervene (see p. 314).

**Tempered Jominy Bars.** The usual Jominy curve relates to the as-quenched condition. To reveal the likeness or difference of two steels after tempering, the end-quenched bars can be tempered alike, that is, with the same time and temperature, and their distance-

hardness curves can then be determined. Bainite tempers more readily than martensite, so the tempered-bar curves may reveal the previous presence of bainite more clearly than did the regular Jominy curve on the untempered bar.

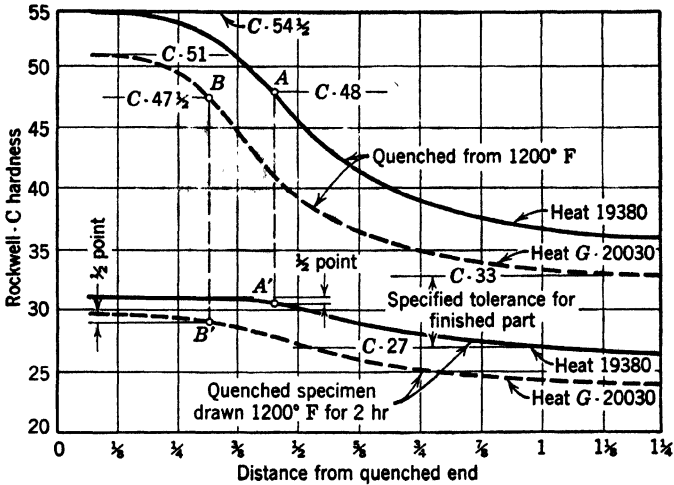


FIG. 176. Jominy end-quench hardness of two heats of 6130 steel. Upper curves: Hardness readings after quenching from 1700°F. Lower curves: Hardness readings repeated after specimens were drawn 2 hr at 1200°F. Shepherd grain sizes: Heat 19380, 9; heat G-20030, 8.5.

Liedholm<sup>31</sup> (using a 1/2-in.-diameter bar instead of the usual 1-in., because he was dealing with stock less than 1 in. thick) shows Fig. 176. He was primarily interested in the tempered hardness. The hardness levels are as expected from the C contents, the only major composition differences, as shown in Table 21.

TABLE 21

	Heat 20030	Heat 19380
	%	%
C	0.29	0.35
Mn	0.71	0.67
Si	0.36	0.45
Cr	0.98	0.98
V	0.20	0.23
Fracture grain size	8½	9

If the as-quenched curves are displaced vertically to the same hardness level, they do not exactly coincide, but their differences are

not very obvious. The differences are made more evident in the curves for the tempered condition. The more bainitic region in heat 20030 tempered more readily than the corresponding more martensitic region in heat 19380.

**Cast versus Wrought Steel in the Jominy Test.** One highly practical use of Jominy testing is in the steel mill, when H steels (pp. 8, 280) are being produced to hardenability specification. A cast Jominy bar reflects the behavior of that particular heat of steel (which results from its chemical composition and its own "individuality") well enough to predict the behavior in the heat treater's hands of the wrought steel, rolled or forged from the ingots of that heat. Caine<sup>10</sup> presented experimental evidence, and the steel mills have accumulated a vast amount of data to corroborate the statement. It is, however, questioned by English investigators.<sup>42</sup>

Of course, if each ingot from a heat has an individual addition, such as boron, each ingot may behave differently.

**Uses and Limitations of Jominy Data.** Jominy data, especially those obtained by metallographic methods so that the 100% martensite criterion may be applied and supplemented by data from tempered bars, give much useful information to the heat treater.

To produce useful information the Jominy test needs to be made on steel austenitized in just the same way and quenched at just the same temperature as will be used in actual production. It differs from practice in that it uses a standardized, rather severe, pressure quench, which may be decidedly more drastic than the production quench. To transfer the results so obtained to those that will be secured with the production quench, that is, to convert Jominy distance to hardenable diameters, requires definite knowledge of the relative intensity of the two quenches, a knowledge which is generally rather indefinite. Compare Manning.<sup>40</sup> However, the standardized Jominy test ordinarily places any two lots of steel in their correct order, as far as evading the  $A_{r'}$  transformation goes. It does not give complete quantitative information as to what goes on at  $A_{r''}$ , even in water quenching and still less in oil quenching.

The different responses of different steels to the rate of cooling in the bainite and martensite regions, and the known shortcomings of slack-quenched bainitic steel indicate the need for supplementing the hardenability test by some other source of at least qualitative information about what happens in the bainite and martensite regions. That need is partially met by conventional S curves, still better by a "quenching S curve."

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## CHAPTER 12

### S CURVES

If austenite is transformed at different temperatures and for different times, a plot showing the amount of transformation accomplished, as determined by metallographic examination, gives a "time, temperature, transformation" or TTT curve, which reveals the basic behavior of austenite in its unstable temperature range. Many such curves have a typical S shape and are commonly called "S curves." To appreciate the limitations of such curves, the method by which they are obtained (discussed on p. 36) needs to be repeated.

A certain heat of steel (which may not accurately represent all steels of its general type) is made into very small thin specimens (so that the inside will cool practically as fast as the outside). These are austenitized at some selected time and temperature, resulting in some particular degree of homogenization and some particular grain size (any other homogenization or grain size will alter the results) and some temperature in the austenite region is selected from which to cool.

A molten lead, fused salt, or similar hot bath is held at a selected temperature below  $1335^{\circ}$ , and the tiny specimen is plunged into the hot bath. It is left there for a measured time, perhaps only for a fraction of a second, perhaps for hours, and then plunged into brine. Let us start with a hot bath not much below  $1335^{\circ}$ , say,  $1275^{\circ}$ .

If no austenite transformed in the time allowed in the hot bath, it will all transform in the brine quench, either to all-martensite or to martensite plus bainite (which condition exists being discovered by etching and microscopic examination). With more time, part of the austenite transforms to ferrite plus pearlite, pearlite, or pearlite plus carbide, depending on the composition. The *remaining* austenite is then changed to martensite or martensite plus bainite in the brine quench. With enough time, *all* the austenite transforms at the hot-bath temperature, and *no* martensite or bainite can be formed on the quench. The pearlite is easily distinguished under the microscope, and hardness determinations also reflect its presence.

This series of specimens gives, for this one hot-bath temperature, the time at which the transformation starts (the time delay, nucleation, or incubation period), the time for a selected degree of partial transformation, and the time at which the transformation is entirely completed.

Another such series at a lower hot-bath temperature gives similar data for that temperature, and so on with other series at other temperatures, until the pearlite region has been mapped, through the 1275–900° range. Somewhere in this range there is a temperature at which the minimum time delay or incubation period occurs, short of which austenite is all retained for later transformation, beyond which some austenite is used up.

**Critical Cooling Rates.** This minimum time delay and the temperature at which it occurs are the basic facts that attend the quenching operation. Any cooling rate which, at that temperature, takes more than the minimum time delay, will result in formation of some pearlite and in failure to get all-martensite on further cooling. The dividing line is called the "critical cooling rate." A *faster* cooling rate at that temperature holds the austenite ready for transformation to martensite at a lower temperature ( $M_s$ ), and, if a suitable cooling rate (not necessarily the same as for avoidance of pearlite) is provided at the lower temperatures, martensite can be formed and bainite avoided.

Plotting temperature against time of beginning of the transformation as found by metallographic examination, we have the contour of the curve for starting time. A similar plotting of the temperature versus time for completion does the same for ending time. Time is plotted on a logarithmic scale. Intermediate curves can be plotted for any percentage of completion.

The peak of the pearlite section of the S curve approaches the zero time axis more or less closely. This peak is called the "nose" of the curve.

As the temperature drops below the nose, the steel is stiffer, and the transformation is more and more retarded, the time delay increases.

In the 900–400° region, a further series of specimens, plunged as before into hot baths held at the selected temperatures in this region, and then quenched, shows, first, this greater retardation due to steel stiffness (a "bay" below the "nose"), and then the time-delay decrease again, the urge to form bainite becoming so great at the tem-

perature drops and the instability of austenite increases as again to overcome the increasing stiffness in some steels. In some it does not overcome it. Thus, the S curve may swing back to form the lower part of the S, the "chin"; but, as in humans, the chin can stick out or recede.

Throughout the bainite region, if the steel is given sufficient time at constant temperature, *all* the austenite will transform to bainite, but in some alloy steels it may take weeks to do so. The steel is so stiff that, although some transformation may begin promptly, completing the process is a lengthy matter, and the last traces of austenite are persistent.

If cooling is fast enough through the bay to preserve austenite when temperature  $M_s$  is reached, martensite starts to form. This temperature may be as high as  $600^\circ$  in low-C steels, but in many steels for quenching it is  $400^\circ$  or below. The  $M_s$  and  $M_f$  points are important in the differentiation of alloy steels as to quenching behavior. At the  $M_s$  point we find a quite different state of affairs from that at higher temperatures, consistent with the fact that instead of pearlite or bainite being produced, each consisting of cubic ferrite plus carbide, there is now being produced tetragonal martensite, quite a different breed of cats.

If the temperature is held a little *below* the  $M_s$  point, only a little martensite will form, and some of the remaining austenite will transform to bainite. To prevent bainite formation, rapid cooling must be kept up.

**Types of S Curves.** Morral<sup>1</sup> has classified the curves into five types. Types 1, 2, and 3 each have three subclasses, type 4 has two, and type 5, none. Figure 177 shows schematic plots of his types

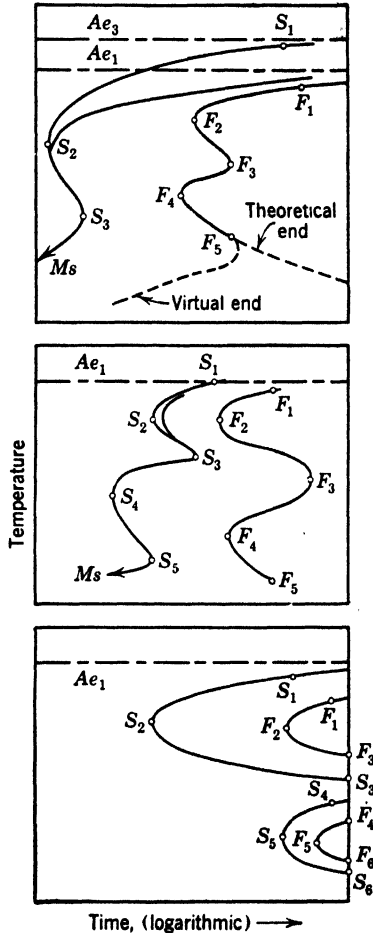


Fig. 177. Some types of S curves.

2, 3, and 5, to illustrate how different they may be. Morral has tabulated nearly 200 sets of data giving the temperature, time, and hardness for selected positions on the curves shown by the letters in Fig. 177. From these tabulated data, the isothermal S curves can be reconstructed with reasonable accuracy on scaling such as in Fig. 152.

Specific steels of the types falling into these classes are given in Table 22.

TABLE 22

Composition								Type	Time Delay	Position $S_5$
C	Mn	Si	Ni	Cr	Mo	W	V		Position $S_2$ at Pearlite Nose	at Bainite Chin
0.38	0.37	.....	.....	0.57	.....	.....	.....	II	0.6 sec.	.....
0.28	0.69	0.41	3.30	0.12	0.03	.....	.....	III	2 sec.	1 sec.
0.83	0.32	0.21	.....	3.89	4.30	5.79	1.30	V	18 min.	29 min.

The temperatures for start and finish of martensite formation also vary widely.

The vast differences in the behavior of different steels as to isothermal transformation make the S curves very revealing.

**Martensite Formation.** We recall that the separation of martensite is practically instantaneous, but only just so much forms at a given temperature. Holding at that temperature produces practically no more martensite. But the next increment of cooling produces more martensite. Martensite forms on a falling temperature. This is probably connected with the pressure or the strain caused by the volume change, for martensite is about 5% less dense, that is, occupies 5% more space than the austenite from which it formed. Martensite starts to form at various points and grows in needles. Soon a network of needles exists, with spots of untransformed austenite within the net. Since the martensite occupies more space and is less dense than the austenite from which it came, the expansion exerts pressure on the remaining austenite.

As Scott<sup>2</sup> puts it, pressure lowers the transformation temperature of austenite to martensite. Manning<sup>12</sup> also discusses this topic. This stabilizing effect of pressure upon austenite or the resulting plastic deformation at a given temperature is the source of some peculiarities in the retention of austenite and accounts for the fact that the temperature must be further reduced in order to form more martensite.

As a consequence of the falling-temperature behavior, if the steel is sluggish enough to need to go below room temperature before all the austenite transforms, "cold treatment" may be required.

The lower part of the S diagram should carry a field for martensite formation, such as the plot given by Cohen,<sup>3</sup> Fig. 178, for eutectoid steel.

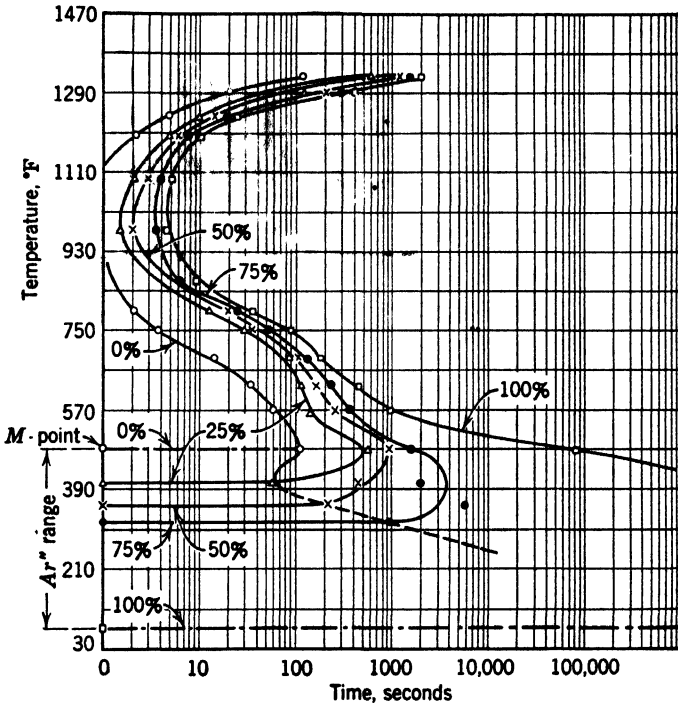


FIG. 178. S curve for eutectoid steel.

toitd steel. The chin of the S curve, showing the time for initiation of bainite formation, is often poorly defined, and the *Ms*–*Mf* range was not shown on the early S curves for many commercial steels. This lack has been largely made up by more recent data by Grange and Stewart.<sup>4</sup>

**Applicability of S Curves.** The "isothermal" technique, holding austenite at "constant subcritical temperature," evaluating the time delay before starting and the time required for completion of the transformations, with the resulting data plotted as S curves, provides a starting point for a crude appraisal of the possibilities and limitations in control of cooling rates. If we want to soften the steel

fully by annealing, we look at the *ending* of the S curve in the upper pearlite region. If we want to control the fineness of the structure, as we may in normalizing (Chapter 17), we are interested in the starting and ending times in the lower pearlite region. Should we want to produce bainite, we are interested in the starting, and particularly

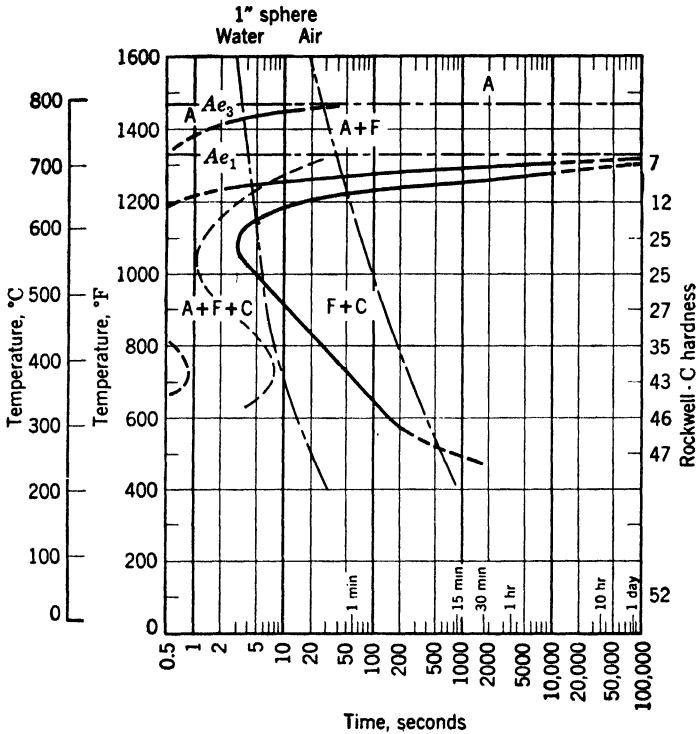


FIG. 179. 1035 C steel. Analysis: 0.35 C, 0.37% Mn. Austenitized at 1500°F. Grain size: 75% 2-3 + 25% 7-8.

the ending times, at the nose and in the bainite temperature region.

When we want to produce martensite we are vastly interested in the time delay in the pearlite range, also in the starting time in the bainite range, and the  $M_s$  and  $M_f$  points.

The austenite stability, as shown by pearlite time delay, bainite time delay and completion time, and the  $M_s$  temperature, is vastly influenced by the chemical composition. However, once the cooling rate is adjusted so that a particular structure is produced, be it, for example, bainite or martensite, the properties of like structures are closely alike. That particular structure of pearlite, bainite or mar-

tensite is in turn greatly influenced by the amount of C, but relatively little affected by the usual alloy content.

Alloy content is provided primarily as a means of producing just sufficient sluggishness of austenite to allow getting the desired structure at the cooling rate imposed by the size of the piece to be treated.

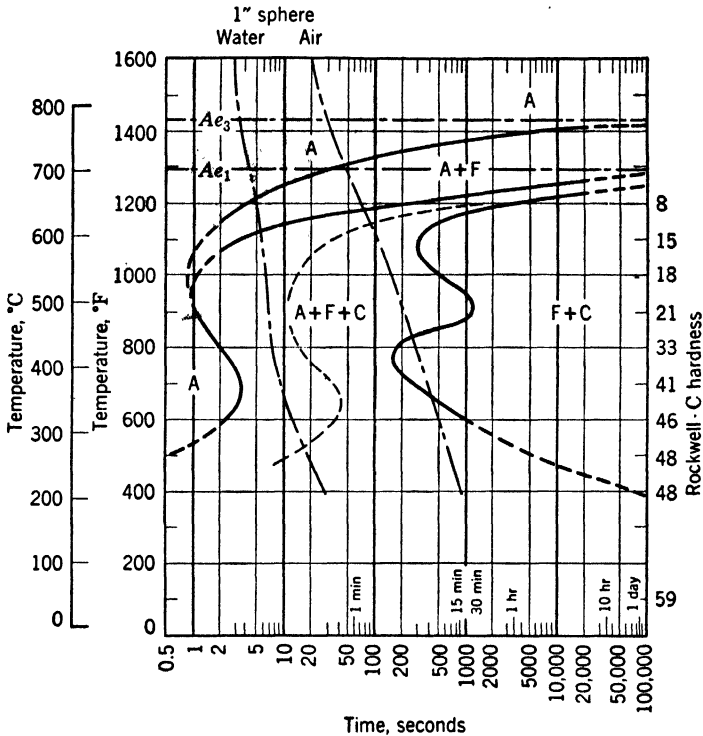


FIG. 180. 1335. Analysis: 0.35 C, 1.85% Mn. Austenitized at 1550°F. Grain size: 70% 7 + 30% 2.

**Influence of Composition.** How alloy and C content influence the S curves is exemplified in Figs. 179–183. Note that time is plotted on a logarithmic basis. The plain C 1035 has less than half a second time delay at the pearlite nose, the N.E. 8949 has a minute. In 1335, with its Mn content, the pearlite time delay is slightly increased, and the bainite chin does not protrude beyond the nose. The N.E. 8949 chin does protrude.

**How Not to Use S Curves.** If we drew upon the conventional isothermal S curves the actual air and water cooling curves for the center of a 1-in.-diameter sphere, as has been done in Figs. 179–



181, then Fig. 179 would indicate that, even on water quenching, the plain 0.35% C steel will fully transform to ferrite and pearlite at the center of the 1-in. sphere but will be harder (of a finer structure) than when air-cooled.

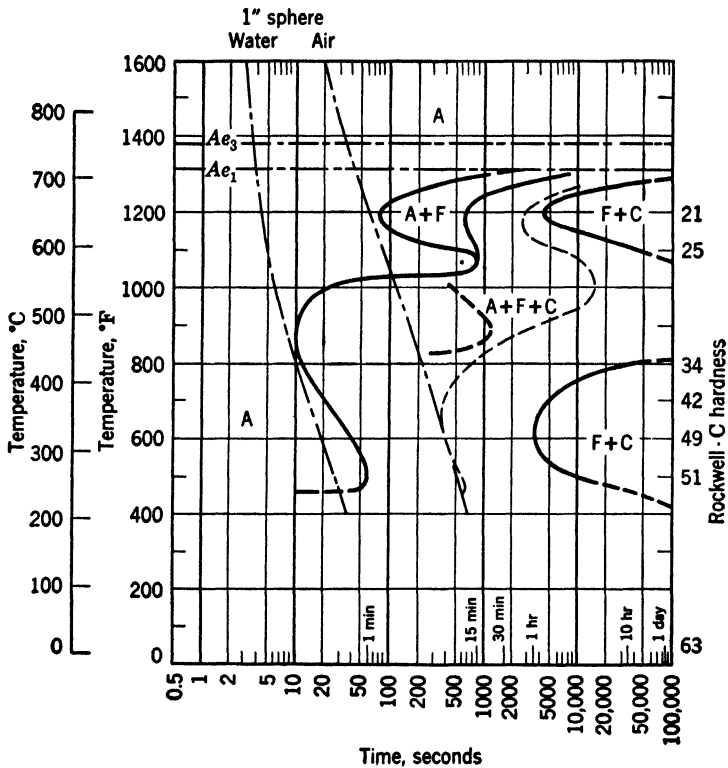


Fig. 181. 8949. Analysis: 0.52 C, 1.19 Mn, 0.51 Cr, 0.53 Ni, 0.35% Mo. Austenitized at 1550°F. Grain size: 7-8.

At face value Fig. 180 says that in 1335 the center of the sphere, on air cooling, will separate ferrite and pearlite from 1300° down to 825° and that the transformation will have been completed, since the cooling curve reaches past the ending of the upper-bainite reaction. It likewise says that, even in water quenching, the center would have transformed some 30% of its austenite to ferrite, pearlite, and upper and lower bainite before reaching 600°, which we may take as the  $M_s$  point.

Similarly, for 8949, Fig. 181 says that in air cooling the center will miss the pearlite nose and will start at 1025° to transform to upper

bainite, but that at the  $M_s$  point, say,  $450^\circ$ , about half of the austenite will have become bainite, leaving half to transform to martensite on further air cooling. In water quenching, the center of the 1-in. sphere easily misses the pearlite nose and just misses the bainite

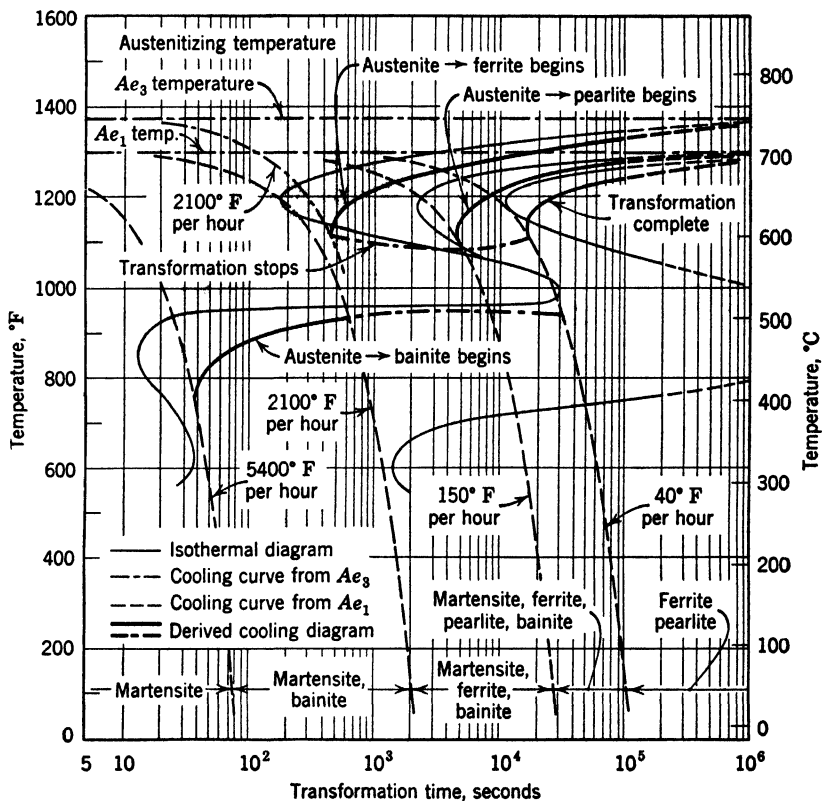


FIG. 182. Attempted correction of isothermal to continuous cooling conditions for 4340.

chin. Thus, in water quenching, the diagram says that the 1-in. sphere of 8949 will completely harden to martensite, but that no larger sphere would so harden. Actually this is *not* true; 8949 is a decidedly deep-hardening steel. A well-austenitized piece so large that it takes 30 sec to cool to  $800^\circ$  at the center, would actually harden to all-martensite at the center instead of some 70%.

The inability of the conventional S curve to represent truly the behavior on continuous rapid cooling is due to the way the S curve is determined.

The constant "subcritical temperature" technique of getting the S curve holds the piece at constant temperature and gives the austenite a chance to transform. If we keep on cooling the piece, we help it to slide past the nose of the S curve, much as cooling an egg

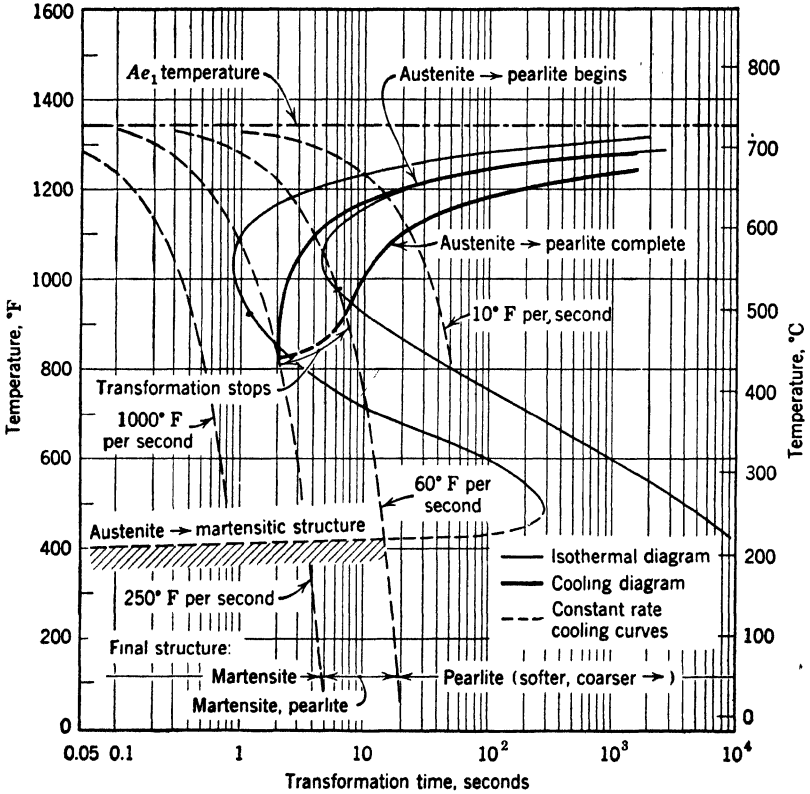


Fig. 183. Diagram showing austenite transformation on cooling at constant rate in a plain-C eutectoid steel. Derived from the isothermal diagram.

about to hatch will keep it from hatching. Hence, in quenching practice, with continued cooling, the pearlite nose's time delay is increased and its temperature lowered, beyond the point found on the S curve.

We need a *different* S curve to evaluate hardenability. Hoyt<sup>5</sup> found that the uncorrected S curve did not lead to true results on some welding problems.

**Attempts at Mathematical Correction.** Attempts to work out empirical corrections have been made by Aborn,<sup>6</sup> Grange and Kiefer,<sup>7</sup>

Gillmor,<sup>8</sup> and others. Grange and Kiefer give "derived" or "corrected" S curves for 4340 in Fig. 182 and for eutectoid C steel in Fig. 183, the light lines being the isothermal S curve and the heavy ones the S curve, mathematically calculated on the basis of *constant* cooling. The basis for this "correction" is not sound, since the rate of cooling, in practice, is *never* constant; it is constantly decreasing as the temperature falls, so that attempts at mathematical correction are only approximations, though possibly useful rough approximations if their accuracy is not overestimated. Dube and Gerstman<sup>9</sup> show a similar "corrected" curve for another complex steel of higher-Cr content. These *calculated* plots are only qualitatively correct, but the effect is in the direction shown.

The *isothermal* S curve is pessimistic when looked at from the quenching point of view.

**Qualitative Effects.** As Aborn puts it, continuous cooling, as contrasted with isothermal holding, shoves the S curve downward some 100°; it also shoves it rightward, in fine-grained material, by some 50% added to the nose time delay, and by some 200% added to it in coarse-grained material. Note the round figures; no such generalization is accurate. He was talking about C steels for welding. (This behavior is not unexpected when we recall that even relatively slow cooling lowers  $A_c$  from the  $A_e$  temperature, that for "infinitely slow" cooling.) The isothermal technique, quenching into a hot bath, gives slower cooling than quenching into a cold bath; hence, the latter pulls down the reaction temperature more than the former. As the reaction temperature drops and the steel gets stiffer, diffusion is less rapid, so that the time increases.

A delay of  $\frac{1}{2}$  sec at 1000°, at the pearlite nose of the S curve for a fine-grained steel, often becomes  $1\frac{1}{2}$  sec at 900° in continuous cooling for a similar but coarse-grained steel. However, even on an accurately corrected curve, few of the C and low-alloy steels commonly heat-treated have as much as 5 sec time delay at the pearlite nose. The highly alloyed steels have a more generous period of grace.

If the data of Fig. 182 are taken at face value, the 180-sec delay at 1200° for start of ferrite deposition becomes 540 sec at 1100°, and the 13 sec at 850° for start of bainite formation becomes 40 sec at 750°. The actual differences in regular quenching will be at least as large. Such differences have real meaning in respect to the depth of martensitic hardening attained on a given quench of a large section.

Let us therefore clearly understand that the isothermal S curve gives only a crude qualitative appraisal of hardenability.

Because of the shortness of the time delay, small differences in cooling rate between outside and inside mean a lot. Only in an extremely thin piece such as a hair spring for a watch can we be close enough to the truth in speaking of a cooling rate. For practical utility we must consider the cooling rates at inside and outside of a cooling piece of steel of larger section.

### COOLING RATES

For the moment, we will lay aside the S curves, and the *response* of cooling steel to different cooling rates and consider merely the *rate* at which steel will cool. One can create heat *within* a piece of steel, by direct electrical resistance or by induction heating, but we do not know how to create cold within the piece. All we can do is put the hot steel in a colder environment and let the heat drain out at whatever rate it will. And, no matter how fast we cool the surface, the heat will not flow out from the center any faster than the thermal conductivity of the steel permits. This rather low conductivity is a bottleneck. All the heat-treatable steels have about the same thermal resistivity and specific gravity, the two factors affecting heat flow. Changes in percentage of C and in normal alloy content make very little difference, and so we can discuss cooling curves of a piece of heat-treatable steel without much reference to chemical composition. The only appreciable variation in the cooling curves arises from the heat evolutions at  $Ar_1$ ,  $Ar'$ ,  $Ar''$ , and  $Ms$ , which lead to slight jogs in the cooling curves at temperatures controlled by the C and alloy content. These jogs are inconsequential from the point of view of the present discussion.

**Air Cooling.** If it is remembered that a 5-sec pearlite-nose time delay in the 1200–900° region is more than most C and many low-alloy steels will stand without transforming some of the austenite, and that in most of these transformation is complete in 10 sec to 1 min, it may be seen that air cooling of ordinary C steels, in ordinary sections, will produce only the ferrite-pearlite transformation, with no chance of martensitic hardening. This was made clear by the classic work of French and coworkers, summarized by French.<sup>10</sup> He made considerable use of spheres in determining surface and center-cooling rates, since the sphere has the simplest geometry of any solid

and will cool equally in all directions. Figure 184 shows data from French for air cooling of spheres.

**Oil Quenching.** In air cooling, the loss of heat from the surface is so slow that the center temperature keeps pace with it to a considerable extent, as Fig. 189 shows. In oil quenching, Fig. 185, however, with its more rapid surface heat loss, the center fails to keep pace so that, to oil-harden to the center of a 2-in. section, a steel must have

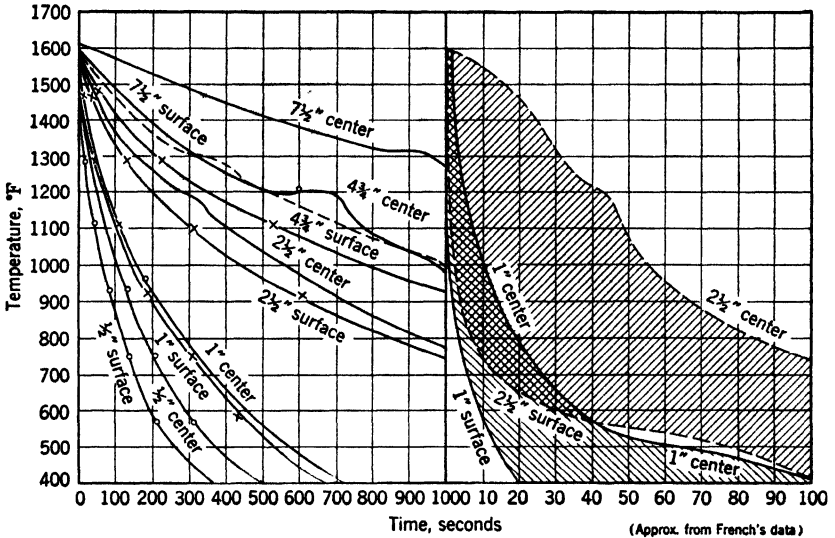


FIG. 184.

FIG. 185.

Figs. 184 and 185. Air cooling (left) and oil cooling (right) of spheres.

around a minute time delay at 950°, and that is a very sluggish steel, requiring high-alloy content. Even in 1-in. section, a time delay of some 15 sec is needed if the center is to harden. In both sizes the surface comes down to 950° in some 3 sec, and so there will be a hardened layer on the surface of many of the common steels even in large sections.

**Water Quenching.** By water quenching, Figs. 186a and b, the surface of even a large section will be reduced to 950° in less than half a second, sufficient to produce a hardened skin on many more of the common steels. But, even in as small a size as 1 in., the center refuses to cool to 950° within 5 sec, and at around 2 in., a steel must have some 25 sec time delay if such a size is to harden at the center. In small sizes one can gain something at 950° on the center-cooling time by using brine instead of water, or by using a high-pressure

water spray instead of still water (Fig. 187). But, from Fig. 186 it is plain that gaining a few seconds at the *center* of a 2-in. section will

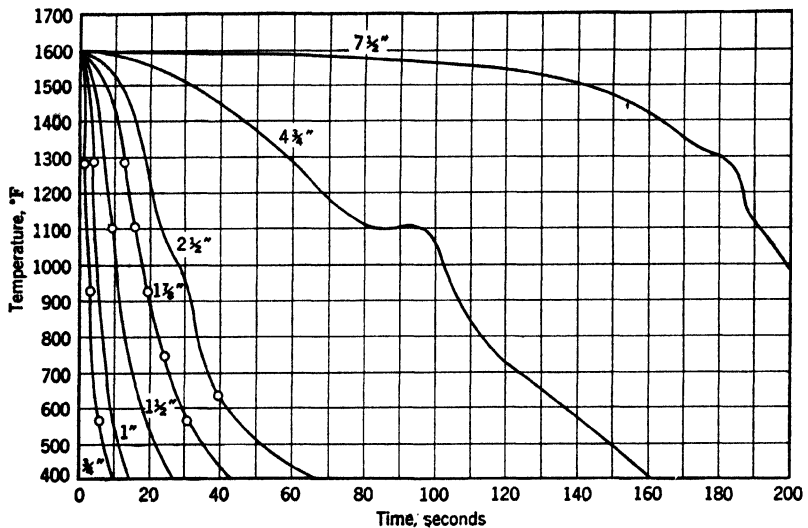


Fig. 186a. Center cooling curves for ordinary water cooling of spheres.

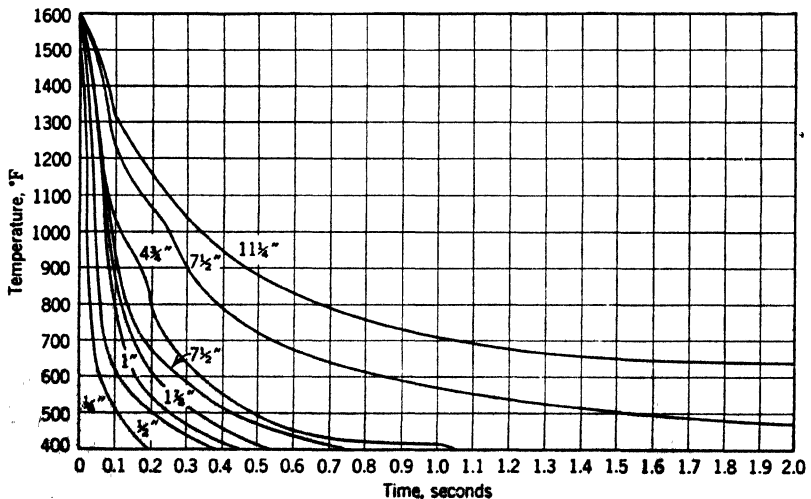


Fig. 186b. Surface cooling curves for ordinary water cooling of spheres.

not be enough. Even with pressure-water or brine quenching, the steel must be quite sluggish to fit the inexorable deliberation with which the center of such a section cools, no matter how fast heat is

extracted from the surface. The behavior of various coolants is discussed further in Vol. II.

**Full versus Slack Quenching.** If we succeed in missing the pearlite nose, we still have to consider whether the center-cooling curve will miss the bainite or chin portion of the true S curve and hit the martensite region, giving a full quench, or cut the bainite curve and give a slack quench.

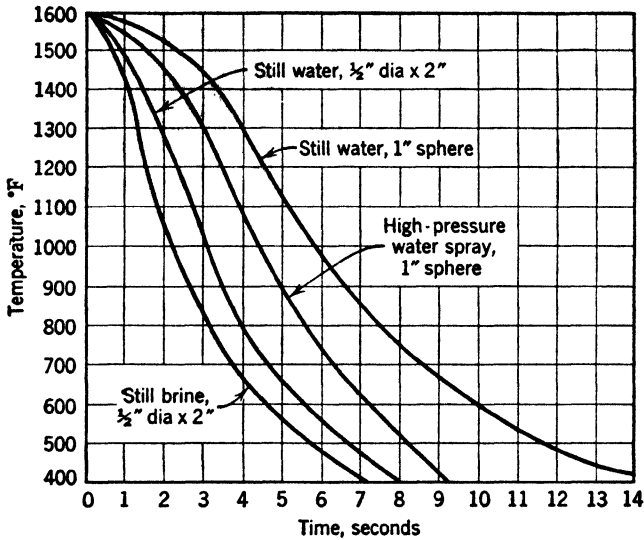


FIG. 187. Center cooling in aqueous coolants.

If we had a true quenching S curve for 4340, that is, one determined with 2½-in. spheres under continuous cooling, and plotted on it the surface and center-cooling curves for air-, water-, and oil-quenched 2½-in. spheres, we would find that the air-cooled specimen has had time to change completely to pearlite. The inside of the oil-quenched sphere has time to transform to a high percentage of bainite, and, since the S curve turns leftward at about 625°, even the surface would contain appreciable bainite before it gets down to 500 or 525°, the *M<sub>s</sub>* temperature for 4340.

The outside of the water-cooled sphere will quench to martensite. The center, however, will contain some bainite and will be slack-quenched.

It is obvious that if we knew, for any steel, any heat of that steel, and any condition of austenitization, and grain size, the exact cooling S curve or the exact correction for continuous cooling as com-



pared with isothermal reaction, and if we also knew the cooling-curve band for the particular section being cooled, for the particular temperature from which cooling begins, therefrom plotting the cooling band across the corrected S curve, the hardenability would be more adequately defined.

**Jominy Bars for Determination of True S Curves.** Liedholm<sup>11</sup> derived a quenching S curve through a modification of the Jominy

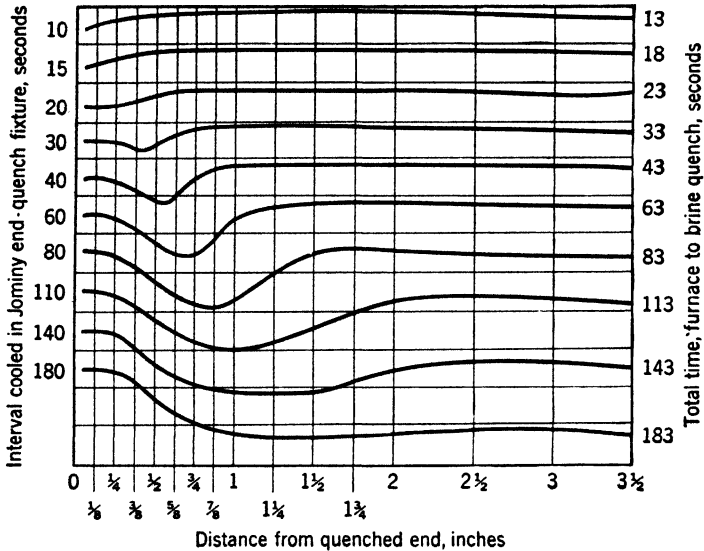


FIG. 188. Hardness surveys on representative end-quench bars of heat 19380; Brine quenched after indicated time in spray fixture. (3 sec are needed to transfer sample from furnace to Jominy fixture.)

technique. Instead of end-quenching the bar with pressure water for a full 3 min, he turned on the water for 10, 20, 30, 40, 60, 80, etc. sec in separate tests, and, as soon as the bar had been quenched for the particular number of seconds, it was dropped into iced brine. This was to determine the time delay the steel would stand. The family of curves for hardness distance from the initially quenched end came out as in Fig. 188.

These curves show that this heat will stand around 20 sec delay before the curve starts to sag, but that from 30 to 60 sec the portion at 1/4 to 3/4 in. from the quenched end, which has been cooled down into the pearlite-nose temperature range on the S curve, by the draining away of heat through the quenched end, has partly transformed:

whereas the air-cooled end is still wholly austenite, that is, not down to the pearlite nose, and responds fully to the brine quench. Similar tests on a less hardenable heat, 20030 (refer to Fig. 176) showed that

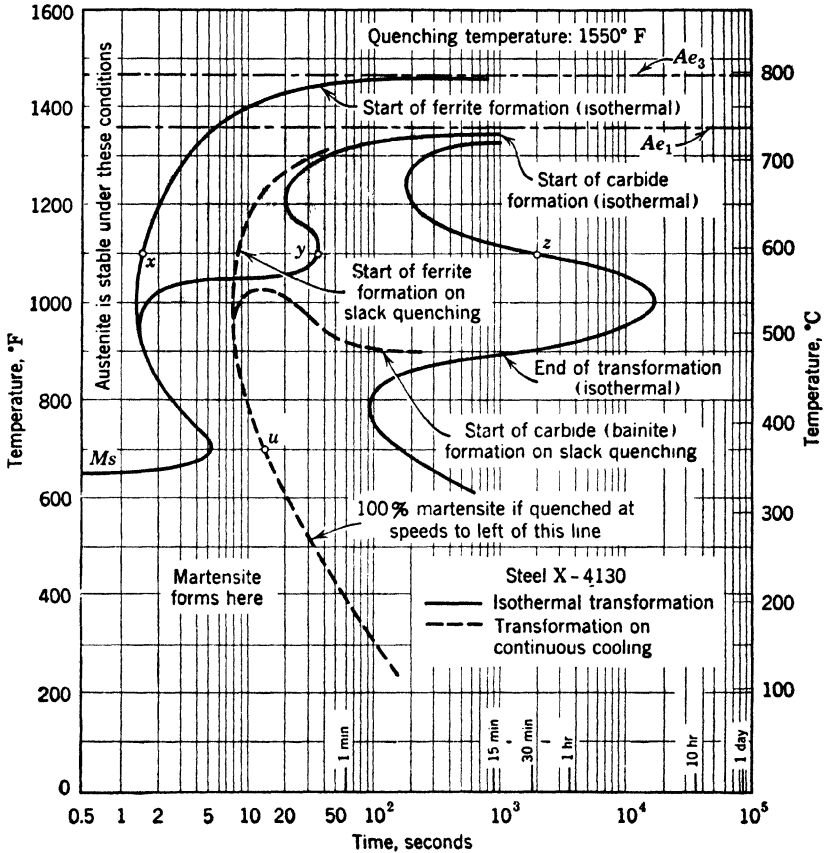


FIG. 189. Transformation curves for SAE X-4130. Full lines are isothermal transformation (from U. S. Steel Corp.'s *Atlas of Isothermal Transformation Diagrams*). Dotted lines are for transformation during continuous cooling as in oil or water quenches. (Liedholm)

its time delay to avoid the pearlite transformation was some 10 sec shorter than with heat No. 19380. By examining the metallographic structures at various positions on these bars, Liedholm constructed an S-curve diagram for continuous cooling, showing the variation in location of the pearlite nose and the bainite curve for each of the two steels. One of his curves is given as Fig. 189.

The differences are of moment and illustrate that a conventional S curve comes far from the truth for continuous cooling. Liedholm has given a similar plot for 4315.

**Another Attempt at Correction.** Manning and Lorig<sup>12</sup> applied the Liedholm method to a series of steels of about 0.30 C, 0.50 Mn, 0.40 Mo, and Cr varying from 0.50 to 2.40%, and examined these interrupted-quench Jominy specimens under the microscope. They measured the distance at which transformation has started during the end quench for a given time of end quenching. When the cooling rate at that position and the time elapsed in quenching are known, the temperature corresponding to this starting of transformation at that point is calculated.

Continuous cooling differs from isothermal holding in that the time spent in the time-delay zone, when nucleation is taking place in the austenite, is spread over a range of temperatures. From the ordinary S curve, it is evident that nucleation is slow above the pearlite nose, rapid at the nose, and slow again below the nose. If a fundamentally correct mathematical relation could be found between the time spent in each decrement of temperature and the degree of nucleation thereby effected, a method of correction of the isothermal S curve into a true S curve for continuous cooling would result. If the "corrected" S curve is really correct, it will give the temperature at which transformation starts at a known rate of cooling, that is, at the position in the Jominy bar which has that rate. Thus a cross-check is obtained on the accuracy of the formula assumed for correction of the isothermal S curve.

Manning and Lorig picked a formula to express degree of nucleation versus cooling rate and made the cross-check on their Cr steels. Fair agreement was found up to 1.5% Cr, but not on the 2 and 2.40% Cr steels. Although the idea that nucleation rate will vary with different cooling rates seems sound, the particular mathematical expression chosen is evidently not universally valid, and so another attempt at mathematical "correction" of the isothermal S curve falls by the wayside, as Grange and Kiefer's did.

Parenthetically, we may note that in all the isothermally transformed Cr steels from  $A_{e_3}$  to about 1200°, the separated ferrite was equiaxed, or "blocky," such as in Fig. 23. At lower temperatures, 1200–930°, ferrite in the form of spears appeared, as in Fig. 24, along with the blocky ferrite, the amount of blocky decreasing and that of spear increasing as the temperature decreased. Below 930° carbide accompanied the spear ferrite to give upper bainite, as in Fig. 26.

**Dilatometric Methods.** By quenching a dilatometer together with its specimen and using special means for recording the rapidly changing dimensions of the specimen, Cook and coworkers<sup>13</sup> determined a cooling S curve for a steel of 0.30 C, 1.63 Mn, 0.49 Si, 0.44 Cr, and 0.33% Mo. Similar technique was applied by Christenson and coworkers<sup>15</sup> to 1025, 1315, 4130, 4340, and 8630. In this case the quenching was by gas.

Cook and coworkers also provided a cooling curve for the center of a water-quenched 1 $\frac{7}{8}$ -in.-thick plate, as shown in Fig. 175, which (A) has been drawn upon the cooling S curve in Fig. 190. By the time that location gets down to 700°, it is 50% ferrite and pearlite. At 675°, the remaining austenite starts to transform to martensite, but at 600° a little bainite begins to form. At 325° the austenite is all gone, but the center location is slack-quenched, containing not only ferrite and pearlite in large amount, but also some bainite along with the martensite. For completely martensitic quenching at the center location, a curve such as B for the center of a 1-in.-diameter water-quenched sphere would be required.

The correlation of cooling S curves and center-cooling curves in Fig. 190 gives a correct picture in contrast to the incorrect one of Figs. 179–181, where *isothermal* S curves were shown.

**Further Limitations of S Curves.** It has been assumed that the right-hand curve of the isothermal S-curve diagram, that for completion of transformation, is reliable for selecting annealing temperature and time. Indeed, on this basis a shift from the conditions previously used to those indicated by the S curve has led to much-shortened annealing schedules.

However, such use of the S curve involves the assumption that a massive section, in the austenitic condition, “quenched” into a hot molten bath, or put within a furnace operating at the desired temperature, in the bay of the S curve where completion occurs most rapidly, will behave just as does the tiny section used in determining the S curve. Considering the center cooling curves of massive sections it is obvious that this should not be the case, since the cooling of the center is so slow that its cooling curve must cut the S curve at a higher temperature, and, in medium-C hypoeutectoid steels, some separation of ferrite and pearlite must occur. The austenite thus becomes richer in C. Thus if this temperature be low, for example, one aimed to produce austempering to bainite and if the steel being used is, say, 4340 and the section size is great enough, the remaining

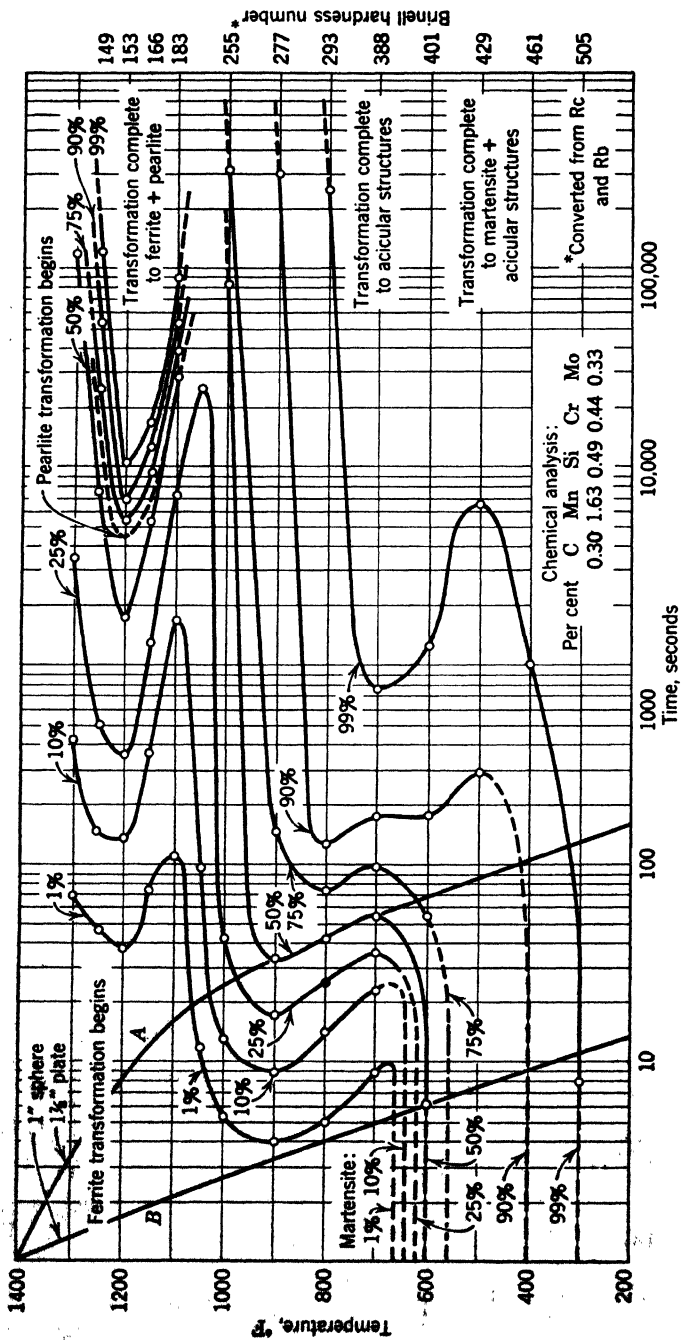


Fig. 190. Transformation characteristics of an alloy steel. Analysis: 0.30 C, 1.63 Mn, 0.49 Si, 0.44 Cr, 0.33% Mo.

austenite may approximate the C content of 4360 when it gets down to the bath or furnace temperature.

The bay in the isothermal S curve for 4340 is at 600° and 30 min for completion whereas that for 4360 is at 540° and 6 hr. The ending curve for 4360 at 600° is at 7 hr. Liedholm and Coons<sup>14</sup> found that ½-in. cubes of 4330, austenitized at 1700° and transferred to a furnace at 600°, were not completely transformed in 6 hr. Post-heating of weldments in 4340, immediately after welding, in the hope of transforming all austenite to bainite and leaving no austenite for later transformation to martensite have shown the inapplicability of the isothermal S-curve data at their face value.

This of course checks with the known fact that austempering is applicable only to thin sections (see Chapter 18).

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## CHAPTER 13

### SO-CALLED CALCULATION OF HARDENABILITY

In the development of the N.E. steels, the aim was to produce steels of graded hardenability with the minimum use of scarce alloy elements. An AISI booklet,<sup>1</sup> published in 1942 contains a chart, Fig. 191, for approximate hardenability levels, calculated on an additive basis, using "composition hardness" factors previously found by the Bethlehem Steel Corporation,<sup>2</sup> for strengthening effects in rolled steels. Within the low-alloy range of the N.E. steels, this was close enough to the truth to allow figuring out such steels, and it places them in substantially correct order of hardenability. Indeed Carapella<sup>23</sup> comes back to a very similar idea. Edson<sup>24</sup> used an additive formula in calculating hardening produced in welding.

However, in 1941, Grossmann<sup>3</sup> made a notable contribution by showing that the relation may be cumulative rather than additive. If two elements are equally effective at a 1% addition of either, using 0.5% of each usually produces deeper hardening than 1% of either alone.

**Hardenability factors.** In the early application of this idea, the criterion of hardenability was taken as the production of 50% martensite. It was also assumed that a given element, whether C or some alloying element, increased hardenability in direct proportion to its percentage, that is, that it had a definite "hardenability factor" such that plotting its contribution to hardenability against its percentage (when no other alloying element is present, or with any combination whatever of other alloying elements) would give a straight line. It was thought that, once these hardenability factors were known, the hardenability of a *complex* steel could be calculated by multiplying the factor for each element by its percentage, and then multiplying these products together. The influence of grain size was recognized and correction factors proposed. The effect of complete versus incomplete austenitization was also recognized.

Modifications of and corrections to the early assumptions came thick and fast. Many experimenters showed that calculations by



the Grossmann factors gave results far from the truth. Modified factors were proposed. Among the articles bringing out the inade-

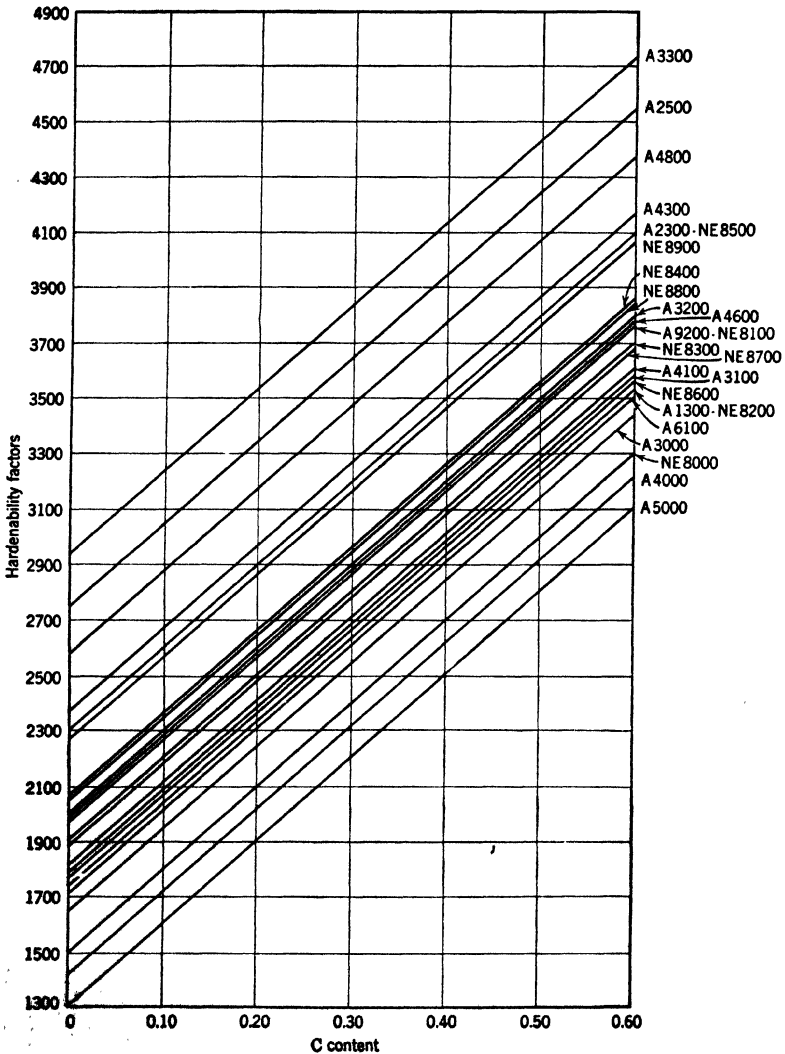


FIG. 191. Approximate hardenability levels.

quacy of the original assumptions are those by Jackson and Luther;<sup>4</sup> Austin, Van Note, and Prater;<sup>5</sup> Ferrall;<sup>6</sup> Kramer, Siegel and Brooks;<sup>7</sup> Steven;<sup>8</sup> Crafts and Lamont;<sup>9</sup> Field;<sup>10</sup> Liedholm;<sup>11</sup> Corbett and Williams;<sup>12</sup> Kramer, Hafner, and Toleman;<sup>13</sup> and Wilson.<sup>14</sup>

Brophy and Miller<sup>15</sup> produced evidence justifying the question whether the Grossmann method is applicable at all. Compare Erb,<sup>22</sup> Schnuecker,<sup>25</sup> and English data of 1946,<sup>28</sup> also.

**Variable Factors.** It finally developed that the contribution to hardenability by an element is *not* a straight-line function of percentage. The multiplying factors are therefore not unchangeable constants, irrespective of the company the element is keeping, but variable. The variables change as the composition varies of the "base steel" to which the addition is made.

**All-Martensite versus 50% Martensite.** Moreover, the criterion of hardenability as a structure containing only 50% martensite, that is, badly slack-quenched, is steadily giving way to the understanding that the proper base line for comparison of hardenable alloy steels is an all-martensite structure.

Likewise, it is becoming understood that it is structure, determined metallographically, rather than guessed at by mere hardness, which is the true criterion.

Appreciation of these matters was greatly advanced by the work of Hodge and Orchoski,<sup>16, 17</sup> who showed the relationship between all-martensitic and partially martensitic structures in terms of hardness and of ideal diameter; revised earlier curves for correlation of Jominy distance to ideal diameter; and gave average curves (not straight lines) for the hardenability factors of Mn, Si, Ni, and Mo. Deviations from these averages for all-martensite and part martensite are also shown. By calculation from several series of steels, they derived a relation between the percentage of C and the "ideal diameter" of unalloyed steel, that is, of pure FeC alloys. This is the basic value on which the calculation of hardenability rests, since the Grossmann formula starts with the  $D_I$  for the C content of an alloy steel and multiplies that successively by the products of the factors for each hardening element times its percentage. Their values differ from Grossmann's; for example, the basic  $D_I$  (for 50% martensite) for 0.50% C steel of grain size 8 taken by Grossmann was 0.22 in., whereas the corresponding Hodge and Orchoski figure is 0.73 in. The adoption of this different base line, plus the use of variable rather than constant hardenability factors for the alloying elements, leads to "calculated hardenability," obtained by a method similar in principle to that enunciated by Grossmann, but quite different in the factors used. Corrections for differences in grain size are applied in either case.

**Haziness of the Literature.** As a result of the continuous stream of modifications and corrections to the original hypothesis, much of the hardenability literature of 1941 to 1946 is difficult of interpretation. The 50% martensite criterion was almost universally used until the end of the period and still persists in some degree. The 50% position on the Jominy bar was usually guessed at by etching, or by hardness alone, without microscopic examination, but this fact was seldom clearly stated. In this transition period, where everybody was working along the original assumptions, it was not thought necessary to emphasize the exact conditions of the test. Later readers, accustomed to up-to-date data obtained on the all-martensite basis as determined metallographically, may easily be puzzled or misled when they read the articles of this period. In the future literature on comparisons of actual versus calculated hardenability, it will be necessary to note whether the Hodge and Orehoski factors, the original Grossmann factors, or still other factors are used.

Hodge and Orehoski point out that experimental data based on the 50% martensite criterion, tend to place the hardenability effects of the various elements in much the same order as do those obtained on the more accurate basis. Hence the early data have a pseudoquantitative and certainly a qualitative value. The interim calculations, however, vastly overestimated the true hardenability.

The 1946 data of Hodge and Orehoski, being obtained metallographically and giving primary consideration to the all-martensite criterion, seem to have a sound basis, and definitely prove that the hardenability influence is not a straight-line function against percentage. Yet an article by Hollomon and Jaffe<sup>18</sup> comes up with some of the old Grossmann factors and some new ones, on the assumption of linearity.

**Bainitic Hardenability.** As already discussed, the concept of ideal diameter has been considered to afford a means for indicating the austenite stabilizing power, or its reciprocal, hardenability, conferred by C or any alloying element. One cannot assume that this will be the same whether the stabilization of austenite is in respect to its transition to pearlite or to its transition to bainite. The S curves for bainite sometimes swing to the right and sometimes to the left, as the martensite formation temperature is approached.

Hollomon and Jaffe<sup>17</sup> sharply differentiate between pearlitic and bainitic hardenability, chiefly on the basis of their belief of the effect of Mo, which, they say, is extremely potent in preventing formation of pearlite but has scarcely any effect on preventing formation of

bainite. They give a plot of "assumed bainitic hardenability of FeC alloys" in terms of  $D_I$  versus % C, for both 50% martensite and all-martensite. Hodge and Orehoski, however, had studied hardenability in relation to percentage of martensite, in steels with quite a range of Ni, Cr, or Mo, in which the other 50% (on the 50% martensite criterion) is bainite. From this, they too give curves of assumed or derived values in terms of  $D_I$  versus % C for 50% and 99.9% martensite, in FeC alloys. The results compare as shown in Table 23.

TABLE 23

$D_I$ for 50% Martensite—50% Bainite	All-Martensite			
	% C	H & J	H & O	H & J
0.10	0.16	0.32	0.09	0.05
0.20	0.23	0.36	0.12	0.105
0.30	0.27	0.49	0.15	0.16
0.40	0.31	0.61	0.17	0.20
0.50	0.35	0.73	0.19	0.24

For the 50% martensite 50% bainite point, Hollomon and Jaffe took the point of inflection of the Jominy curve, plotted on logarithmic paper, as the criterion. Hodge and Orehoski made metallographic observations. Hodge and Orehoski's steels were of 8 to 9 ASTM grain size. Hollomon and Jaffe's grain size is not clearly stated, but apparently was about 7. Hollomon and Jaffe find that grain-size changes have less effect on bainitic than on pearlitic hardenability.

Both pairs of authors proceed to calculate the hardenability factors for industrial elements and, in reverse, to calculate the hardenability in  $D_I$  for complex steels by methods in which the "base-carbon hardenability" is successively multiplied by hardenability factors for each element, times the percentage of the element, using their own factors for C. Although the differences shown in Table 23 do not appear very great for all-martensite, they are appreciable for 50% martensite, and, when they are multiplied by the product of the factors for alloying elements, even the smaller differences become great.

**Calculated versus Observed Hardenability.** Each pair of authors shows plots of observed  $D_I$  versus  $D_I$  calculated by their own system; Hollomon and Jaffe applied their system to the series of steels reported in the earlier publication of Hodge and Orehoski. In that earlier publication these authors used, and Hollomon and Jaffe still use, a single factor of each element, irrespective of its percentage. Hollomon and Jaffe use, as the factors for bainitic hardenability, the same factors for Mn, P, S, Si, and Ni earlier given by Grossmann but

a slightly smaller factor for Cr, and they deny Mo any multiplying factor at all, in respect to bainitic hardenability. By this selection of factors, they come out with calculated versus observed  $D_I$ 's for 50% martensite, 50% bainite, with less scatter than the earlier data of Hodge and Orehoski, but which still show such extremes as 3.0 calculated, 3.6 observed; 4.3 calculated, 3.4 observed; that is deviations of +0.6 to -0.9 in. On the other hand, using factors shown by their 1946 curves, which increase as the percentage increases, and using their own values for base FeC hardenability, Hodge and Orehoski were able to show closer agreement between calculated and observed  $D_I$ 's.

**Mathematics Do Not Lie, but They Must Start with Correct Assumptions.** No basic theoretical reason has been adduced to justify any particular formula for the multiplicative relations; it is merely an observed qualitative fact that the influence of alloying elements is usually cumulative. By choosing suitable constants and suitable factors, several empirical mathematical formulae can often be found to fit fairly closely to an observed set of points. In discussion of Brophy and Miller's paper,<sup>15</sup> Crafts introduced modified factors to make the Brophy and Miller curves fit into the multiplicative scheme.

When you know the answer, you can juggle the method of calculation to give the answer. Unless there is a proved basic relationship to justify the formula, there is little assurance that the formula will fit another set of observed points in any case where you do not know the answer. Hodge and Orehoski are careful to say, for their 1946 data: "This correlation does not necessarily imply that these factors themselves are correct for other alloy combinations."

Kramer, Siegel, and Brooks<sup>7</sup> report that in some cases a combination of two or more alloying elements is *less* effective than would be predicted from their independent behaviors, although the general expectation is that it will be more effective.

Wide divergence is apparent between the assumptions made by Hollomon and Jaffe, and Hodge and Orehoski in respect to Mo. Both pairs of authors are discussing bainitic hardenability. Hollomon and Jaffe used steels with 0.22 and 0.49% Mo, but ascribe a multiplying factor of 1 (no bainitic hardening power at all) to Mo, whereas Hodge and Orehoski give a factor of 2.1 to 0.22% Mo; one of 4.7 to 0.49% Mo.

In a series of steels with 0.40 C, 0.60 Mn, 0.20 Si, and 0.01 to 0.74% Mo, by metallographic examination Hodge and Orehoski found the results shown in Table 24.

TABLE 24

% Mo	<i>For 99.9% Martensite</i>		<i>For 95% Martensite</i>		<i>For 50% Martensite</i>	
	<i>Distance on</i>		<i>Jominy</i>		<i>Jominy</i>	
	<i>Bar, Inches</i>	<i>D<sub>I</sub></i>	<i>Distance</i>	<i>D<sub>I</sub></i>	<i>Distance</i>	<i>D<sub>I</sub></i>
0.01	0.100	0.75				
0.21	0.185	1.35	0.22	1.6	0.39	2.55
0.50	0.600	3.35	0.57	3.25	1.09	4.75
0.74	1.165	4.95				

This seems to be justification for their statement that Mo does influence bainitic hardenability, since the structure accompanying martensite in the 0.21–0.74% Mo steels was bainite. Yet Manning and Lorig<sup>21</sup> found that Mo did not appreciably affect bainitic hardenability.

The Hodge–Orehsoski factor curve for Ni does not check with the data given by Brophy and Miller.

**Possible Use of Hardenability Factors.** One might try to use the relative “hardenability factors” of the different alloying elements as guides were the hardenability factors correctly known for the particular case in hand. However, it is pertinent to point out that, in view of disagreement among authorities as to the behavior of Ni, Cr, and Mo, using such guides may lead to getting lost in the woods instead of coming out on the right trail. As soon as one places reliance on a method of calculation he is likely to use it for prognostication.

An interesting attempt at the use of “calculated hardenability” in prognostication has been developed by Hostetter<sup>19</sup> for finding the lowest-cost Cr–Ni–Mo combination to produce a desired hardenability. The “ideal diameter” and the 50% martensite criteria are used. Including the factor of recovery in steel from the alloy added, it is calculated that the recovered alloys cost about as follows: Cr 16 cents per pound, Mo 89 cents per pound, Ni 31½ cents per pound; that is, for the same cost, one may have 0.11% Cr, 0.02% Mo, or 0.055% Ni.

The amount of new alloy to be added depends on the residual retained from the remelted scrap. Each added increment of a given alloying element is generally less effective than the previous increment, a sort of law diminishing returns. At some point it will pay better to put in a little of a different element than to keep on increasing the same element, but which other element to use depends on its cost and efficiency as compared with another. Using the

original multiplication Grossmann formula and factors, Hostetter figured out the individual increment curves, and a combined Cr-Mo-Ni increment curve. From these he figures the most economical utilization of scrap of various compositions.

Jaffe and Hollomon<sup>20</sup> introduce other requirements in making a calculation similar in principle to that of Hostetter. These requirements are avoidance of quench cracking and of temper brittleness.

Since quench cracking appears the worse, the lower the  $M_s$  temperature, and C strongly lowers  $M_s$ , they advocate keeping C low and using alloying elements to give the necessary hardenability. Since Si confers hardenability but, according to these authors, does not lower  $M_s$ , Si is their first choice. (However, there is a limit to the applicability of Si because of mill difficulties in rolling steels high in Si.)

Jaffe and Hollomon start by figuring *bainitic* hardenability from composition, using the earlier data of Grossmann (not the newer Hodge and Orehsoki data), save for a different factor for Cr. Thus they calculate how much Mn, Ni, Cu, and Cr will be needed to produce a required bainitic hardenability. If the pearlitic hardenability, similarly figured, is insufficient, Mo is included in the list of alloying elements, on the ground that, although they ascribe it no effect on bainitic hardenability, it is accepted by all as very potent in increasing pearlitic hardenability. Moreover, Mo may need to be included to mitigate temper brittleness.

Thus they reach into uncharted regions to concoct a steel especially designed and balanced to meet the hardenability, quench-cracking, and temper-brittleness requirements. They come out with quite definite conclusions as to an optimum range of the content of Mn, Ni, Cu, and Cr. Such conclusions, to be really accurate, obviously depend both on the correctness of the method of calculating hardenability, and on the correctness of the factors chosen to make the calculation with. Such a calculation might be justifiable as a means of selecting a composition with which to experiment. Without experiment, it cannot yet be classed as conclusive, for any such calculations as Hostetter's and Jaffe and Hollomon's are no more reliable than the basic assumptions. Using the Hodge-Orehsoki factors, the Grossmann factors, or the Hollomon-Jaffe factors will give three different results. If we refer to the Brophy-Miller paper, we find evidence that, at least in the case of Ni, none of these three sets of calculations would tell the truth.

**The Boron Situation.** Until the advocates of calculation can agree on basic facts, calculation is a weak reed to lean upon. Even were there agreement on the factors for C, Si, Mn, Ni, Cu, Cr, Mo, etc., the calculation applicable would be upset by the introduction of a trace of B. Boron, present at somewhere around the amazingly low

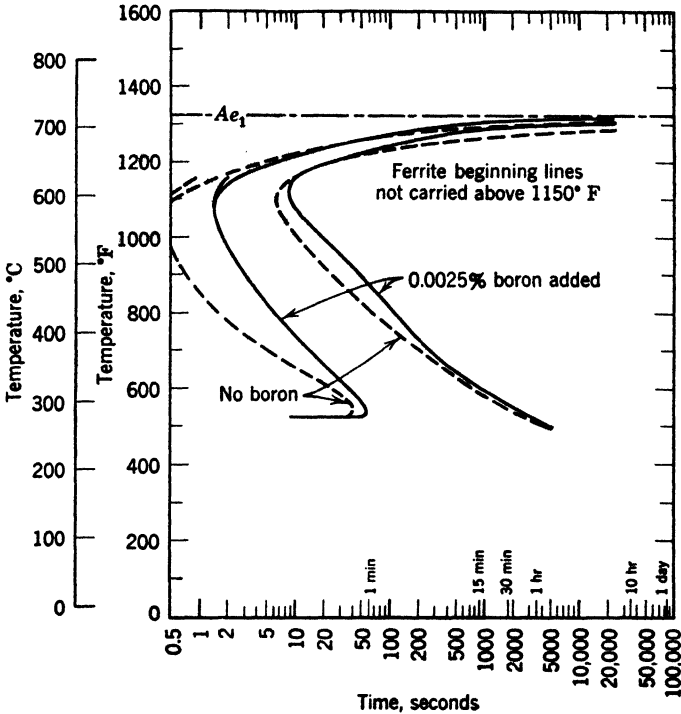


FIG. 192. Comparison of isothermal transformation curves, 0.63 C, 0.87% Mn steel with and without boron. Both austenitized at 1500°F; grain size: 5-6. (From U. S. Steel Corp.'s *Atlas of Isothermal Transformation Diagrams*)

level of 0.0003%, and, having been effectively introduced, increases hardenability in a vast degree (compare Fig. 192).<sup>\*</sup> Its effectiveness is not definitely measured by the percentage found by analysis. Neither chemical nor spectrographic analysis for such small traces of B is of extreme accuracy, in the first place. Moreover, it is established that 0.0002% can be present and be either effective or ineffective. Further, at some not clearly defined percentage, the effective-

<sup>\*</sup> Note that the pearlite nose is shoved back but *not* down, by addition of B. Here is a case where the Grange and Kiefer, or any similar method of "correction" to continuous cooling conditions, simply cannot work.



ness begins to decrease, 0.0005% may have the same or less effect than 0.0003%. The same B addition to different heats, of the same composition as to the elements used in hardenability "calculation," may not produce the same hardenability, and even the same addition to successive ingots cast from the same heat may produce appreciably different hardenabilities. These matters are discussed in Vol. III.

Boron is a cheap intensifier of hardenability, and is plentifully available from domestic ores. It will evidently be used increasingly in production of steels to hardenability specifications. If the heat treater is confronted, as he will be more and more, with a simple MnB, or a complex MnNiCrMoVB steel, the presence of B adds material uncertainty to the already uncertain calculation of hardenability from the balance of the composition.

**Methods for Actual Determination.** Hence whatever virtue calculation may have inures to the steel maker rather than the steel user. If the steel maker has a rapid spectrographic analysis of the composition of a heat while it is in the furnace, and from the analysis is able to figure out what alloy additions he must make to bring the heat out to meet the hardenability specification, that ability has value. However, it is an open question whether casting a Jominy bar and making the Jominy test while the heat is in the furnace, a method increasingly resorted to by steel makers, is not as rapid as spectrographic analysis and calculation, as well as more reliable. Moreover, the experienced steel metallurgist can glance at the analysis and, by comparison with other heats whose Jominy behaviors are on record, appraise the probable hardenability by inspection without resorting to calculation.

Since there is a doubt about the hardenability factors of elements so important as Ni, Mo, Cr, and B, since the calculation is for an ideal, not an actual quench, and the correction to the actual quench is not an easy matter, and since corrections have to be made for differences in grain size and for any lack of complete homogenization of austenite, it seems odd that calculation has intrigued so many experimenters when it is easy to make a Jominy test and to make comparisons directly on Jominy distance, rather than on  $D_I$ , since transformation of the Jominy curve to the fully hardened diameter of a piece in the actual quench is likely to result in inaccuracy.

Brown<sup>27</sup> set out to determine how much hardenability was obtained per unit cost in commercial alloy steels. That is not a very logical basis for a choice between steels unless they chance to have the same hardenability, for a piece of a given section for a given use needs a certain degree of hardenability, no less, and there are some objec-

tions or concomitant drawbacks, to supplying excess hardenability. But it is at least an interesting basis of comparison.

He made the comparison on the basis of "calculated hardenability," and then made it again on the basis of actual Jominy data. The two did not always place the steels in the same order. Brown ascribes this to the fact that "some combinations of certain elements in various contents produce totally unreliable calculated hardenability."

**Appraisal of Hardenability.** To sum up, the usual S curves depict behavior under isothermal conditions but not under continuous cooling. No truly sound method of "correction" to continuous cooling conditions is known. True cooling S curves can, however, be established.

"Calculation" of hardenability has no sound basis and is in a state of chaos. We might better be satisfied with a qualitative understanding, and use that as a basis for experimentation than rely on pseudo-quantitative calculation.

Especially in view of the individuality of different heats of steel a Jominy test on the lot the heat treater is going to use is worth much more than any approximation. The Jominy tests, made on two lots, each with the homogenization and grain size to be used in practice, will compare the lots without the necessity of making doubtful corrections.

The Jominy test does not eliminate a correction for intensity of quench, nor does it reproduce the pressure conditions attending the formation of martensite in large sections. To eliminate these uncertainties the actual section and the actual quench must be duplicated. Thus to get results whose reliability is without shadow of doubt, applying the actual conditions and sectioning and examining the piece must be resorted to.<sup>26</sup> This brings us right back where we started with Fig. 147. The short-cuts are useful, but their limitations should not be forgotten!

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## CHAPTER 14

### PRETREATMENT OF AUSTENITE, HOMOGENIZATION, OVERHEATING

Before a steel is ready for those types of heat treatment involving transformations of austenite on cooling, it must be heated so that austenite is formed. But merely producing austenite is not enough. In the discussion of the S and the Jominy hardenability curves, it has been repeatedly commented that the condition of the austenite vastly affects its behavior on cooling, and hence the properties resulting from these different behaviors. These behaviors depend on the degree of homogenization of austenite and on the grain size produced in the austenite.

#### FORMATION OF AUSTENITE

We have commented (p. 274) on the trigger action of ferrite or carbide present with austenite, and on the similar action of "cored" austenite crystals, in accelerating the transformation of austenite. When the cold steel, containing ferrite and pearlite, or pearlite and carbide, is reheated to produce austenite, the areas that were previously pearlite have, in the first case, to lose C by diffusion into the areas that were previously ferrite, and in the second case, to receive C from the initial carbide areas. Even the pearlite itself, consisting of interstratified layers of ferrite and carbide, has to absorb the carbide into the gamma-iron layers that were previously ferrite. This, however, is facilitated by the contiguity of the layers; the distance over which diffusion of C must take place is small.

A similar necessity for diffusion exists if an alloying element has segregated during the initial freezing of the steel from the melt. Banding, a consequence of segregation, has been discussed (p. 80). The difference in hardenability when the carbide of a hypereutectoid steel is or is not completely dissolved is striking.

The initial condition of the cold steel, before austenitizing, governs the ease of homogenization. A cast structure, whether in an ingot or a steel casting, is difficultly homogenized. Breaking up the

coarse ingot structure by hot working makes homogenization of austenite much easier. Steel castings require much higher temperatures and longer times than wrought steel, since the slow-cooled cast structure has coarse pearlite grains surrounded by thick, practically carbon-free ferrite envelopes.

## OVERHEATING

In hot forging, there is a great temptation to make the steel as plastic as possible, for ease in forging, and to heat so rapidly, in order to save heating time, that, by the time the center is at proper forging temperature, the outside has been overheated. Carried to the extreme, this produces "burnt" steel, a term not related to oxidation, but to heating so hot that actual incipient fusion starts at austenite grain boundaries. This makes a very brittle structure and ruins the piece; there is no way of remedying the condition.

When overheating is severe, but short of actual burning, a coarse structure is produced, difficult to refine by heat treatment, since the coarse grains, surrounded by heavy ferrite envelopes, are very reluctant to homogenize.

For example, an 8-in.-diameter 0.40% C steel shaft, which had been overheated, had a very coarse structure much like Fig. 193a. The tensile test showed low ductility due to coarse grain. The tensile bar had a flat fracture.

Small pieces having an initial structure of this nature usually can be refined by a simple anneal, but with these large pieces a double treatment may be required.

The first heating at 1450°, followed by cooling in the furnace, served to break up the coarse grain (but without full refinement). It is evident that this steel has been very severely overheated. Note the spine growth of the excess ferrite.

TABLE 25

## REFINING OF OVERHEATED STEEL, 0.40% C-8 IN. DIAMETER

	(a) <i>Initial, Coarse, Overheated Structure, as Forged</i>	(b) <i>Annealed at 1450°</i>	(c) <i>Annealed at 1450°; Then Spheroidized at 1400°</i>
Tensile	104,000	90,500	60,000
Yield	58,250	50,250	29,000
Elong., %	7	22.5	43
R.A. %	3	37	69



(a) As forged.

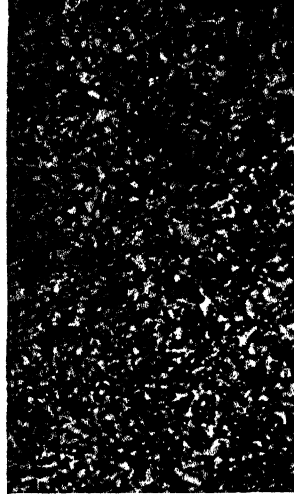
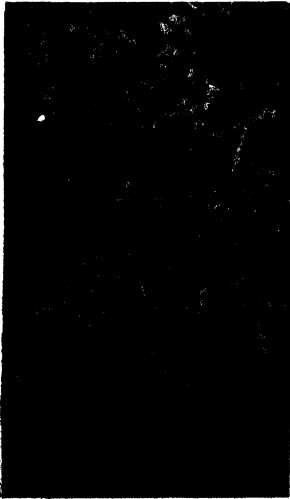
(b) Same, 1 in. bar refined by one long anneal.  
× 100. (Bullens)

FIG. 193. Refining of a small section of severely overheated steel.



(a) After one double treatment, 1450° and 1400°.

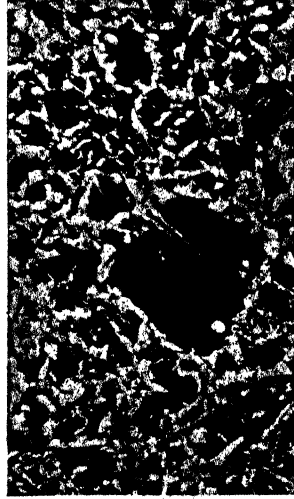
(b) Same, after repetition of the double treatment.  
× 100. (Bullens)

FIG. 194. Difficulty of refining a large section of overheated steel. This was a 10-in.-diameter shaft, with initial structure like that of Fig. 193a.

In small pieces, an unusually thorough soaking at the ordinary annealing temperature will entirely break up the old structure and refine the steel. This is shown by Fig. 193*b*, representing the result of annealing the 1-in. bar of Fig. 193*a*, in the manner previously indicated.

The annealing of large sections with a similar initial structure presents much more difficulty. The result of an ordinary double annealing of a 10-in. shaft, which had an initial structure similar to that of Fig. 193*a*, is shown in Fig. 194*a*. The grain size still is large, although the tensile results indicated good ductility. A second double anneal produced the results indicated in Fig. 194*b*. Although the steel largely is refined, there still is the tendency for large grains or "islands" here and there to remain unaffected.

In such cases, as in steel castings, a double treatment is often required. The first treatment involves a high austenitizing temperature and long time to get an approach to homogenization, even though considerable grain growth of austenite is produced. The piece is then air-cooled, the austenite thus being converted to ferrite and pearlite; the proeutectoid ferrite is better distributed than it was in the first place. When this structure is reaustenitized, with thinner proeutectoid ferrite, the carbide in the pearlite does not have too far to go, and the second austenitizing temperature and time can lower, and one can still get good homogenization without the excessive grain growth of the first austenitization. The austenite, now more homogeneous but fine-grained, is ready for annealing, normalizing, or quenching, to products with better properties than would result from a single treatment.

**Faceted Fractures.** Repetition of austenitizing treatments is among the methods for reclamation of overheated steel showing a peculiar ailment. Apparently some phenomenon is involved in this ailment beyond that of grain growth and deposition of heavy ferrite bands at grain boundaries, and spears along crystallographic planes of the original austenite crystals, since such segregation should yield to a reasonable homogenizing soak. The ordinary burnt or coarse structures are recognizable under the microscope, but some lots of steel which have the usual microscopic structure and would not be suspected on that score, nevertheless show, even in quenched and tempered condition, a fracture containing flat facets where the steel has locally pulled apart along the original austenite grain boundaries instead of fracturing through the grains. The facets are most easily seen in notched-bar impact fractures or fractures of test lugs.



This behavior is common in highly alloyed deep-hardening steels, such as are likely also to show "flakes," but, whereas flakes are commonly considered as ascribable to H, the flat-facet appearance is not ordinarily so ascribed. A fracture with flat facets connotes low ductility, poor notched-bar impact resistance and probably poor fatigue resistance. The facets are quite clearly a result of overheating in forging.

Much attention has been given to the problem of finding a heat treatment to eliminate the facets and restore the normal properties. Observations of the phenomenon in the plant and laboratory and trials of various heat treatments are reported in a series of papers of the British Iron and Steel Institute.<sup>1-5</sup>

The facets were found in steels heated above 2300°, especially with long holding times. Steels of the same nominal composition vary greatly in susceptibility.

In individual cases improvement is brought about by (a) reheating into the actual overheating range, up to 2500°, and slow-cooling to 2300°; (b) repeated renormalizing; (c) 8 hr at 2250°, cool; 8 hr at 2000°, cool; 8 hr at 1750°, cool; (d) holding several hours at 2100°, cool (the normal quench and temper is applied after such treatments); (e) repeating the quench and tempering.

The hardenability may be either decreased or increased from that which is normally characteristic of the steel, when overheating has occurred and the facets are in evidence. The S curve may be displaced, either up or down.

**Variability in Behavior.** The British investigators found C steel less prone to overheating effects than highly alloyed NiCr, or NiCrMo steels. Addition of Al sometimes lowered the overheating temperature; that of V sometimes raised it. Basic electric steel often has a lower overheating temperature than basic open-hearth. Sulphur, nonmetallic inclusions, gases within the metal, and the heating-furnace atmosphere were all studied without clear results save that a steel with large content of nonmetallic inclusions often had a higher overheating temperature than a cleaner steel. That is, in this connection they find some virtue in dirty steel! The furnace atmosphere is not held to be influential. The hypothesis most favored is one involving precipitation, in the overheating range, of some material at the austenite grain boundaries which does not diffuse readily at lower temperatures and which makes the boundary so affected fail to respond to heat treatment in the same way as the body of the grain. No clear picture emerges as to the exact cause

nor the exact remedy, save the avoidance of overheating. There is no agreement as to how to treat a steel once overheated, and so when the heat treater meets such a case, he will have to experiment. Obviously a repetition of the quench and temper will take less time and involve less danger of scaling or decarburization than the other methods, and hence is worth trying first. Similar studies were carried out in the United States by Haworth and Christian<sup>6</sup> on 4340 and by Strohm and Jominy<sup>7</sup> on several alloy steels. Overheating in forging, unless a considerable amount of forging reduction is thereafter accomplished, produces facets in the fracture when the steel is subsequently reheated to a normal hardening temperature, quenched, and tempered. The tempering range may be from 270 to 430 Brinell, without greatly modifying the appearance of the fracture, but around 300 seems to bring the facets out most clearly.

The facets are definitely ascribable to grain coarsening. High temperature in heating for forging or long time at normal temperature will produce it, and some heats are more susceptible than others. Fractures with facets look terrible, but, on quenched and tempered stock, grain-refined by heat treatment, the vestiges of the original coarse austenitic grain size do not appreciably impair the mechanical properties, even when the facets are large and plentiful. However, at the extreme in size and abundance of facets, the ductility and the notched endurance fall off sharply. The dividing line is not clear, and so inspection by appearance of the fracture alone is difficult. Strohm and Jominy found tensile ductility to show impairment of properties, whereas notched-bar impact tests were not so sensitive.

**Hardenability.** It is common experience that in steels containing carbide-forming alloying elements, notably Mo, W, and V, if the same austenitizing times are used, but the temperature increased, at some elevated temperature there will be a notable increase in hardenability. This is shown in the thermal-analysis curves of Fig. 8. Swinden,<sup>8</sup> indeed, found the behavior so marked in W and Mo steels that he termed this temperature "*T* max" and used it as a signpost of behavior. This effect is a mixed one, both homogenization and grain growth being involved.

Roberts and Mehl<sup>9</sup> studied two practically eutectoid steels of like chemical composition save for Al, the low (0.008%) Al steel being easily coarsenable, the high (0.06%) Al difficultly coarsenable. These were equally coarsened, to 6-6½ fracture grain size, the low Al at 1345° for 10 min, which left residual carbide, the high

At 1610° for 30 min, which dissolved the carbide and homogenized the austenite without coarsening the grain.

The upper portion of the S curves, that concerned with normalizing, was determined, with the results shown in Fig. 195. The right-hand pair of curves for homogenized steel, leads to transformation at lower temperature, a longer incubation period, and a shorter time for completion of transformation once it starts, all fea-

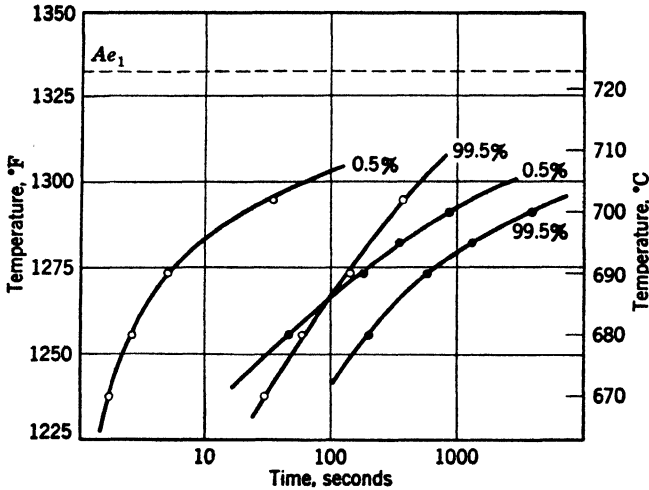


FIG. 195. Effect of undissolved carbide on position of the S curve. Eutectoid steel. Pair of curves to left represents inhomogeneous austenite (steel C). Pair of curves to right represents austenite free from carbide (steel D). Fracture grain size: 6-6½. 0.5% represents beginning; 99.5% represents ending of the transformation.

tures tending to create finer pearlite and a stronger product. This experiment eliminates the factor of grain size. It justifies the practice of austenitizing at high temperature for normalizing.

**Prenormalizing.** That prenormalizing so as to produce fine pearlite and short paths for diffusion of carbide in austenitizing favors hardenability on quenching is also indicated by the work of Roberts and Mehl, but that effect is considered to be a minor one as compared to that of grain size.

Working with a series of steels, all resistant to coarsening so that, irrespective of the austenitizing temperature used, the grain sizes were always alike in each steel, Welchner, Rowland, and Urben<sup>10</sup> studied the hardenability in the Jominy end-quench test of steels with various structures prior to austenitizing for the hardenability

test. The initial Brinell hardnesses of these structures in 1¼-in. bars are shown in Table 26.

TABLE 26

	4620	9442	4140	3240	4340
Commercially spheroidized	131	156	152	182	171
Annealed to mixed lamellar and spheroidized structure for machinability	*	170	170	187	179
Hot-rolled	179	262	331	302	285
Normalized from 1750°	187	255	277	331	311

\* 4260 normalized from 1600°, 170.

Obviously the pearlite of normalized 9442 is fine, from the hardness shown in the Jominy test, as indicated by Fig. 196. The hardenability curve was the same whether quenched from 1450° or at

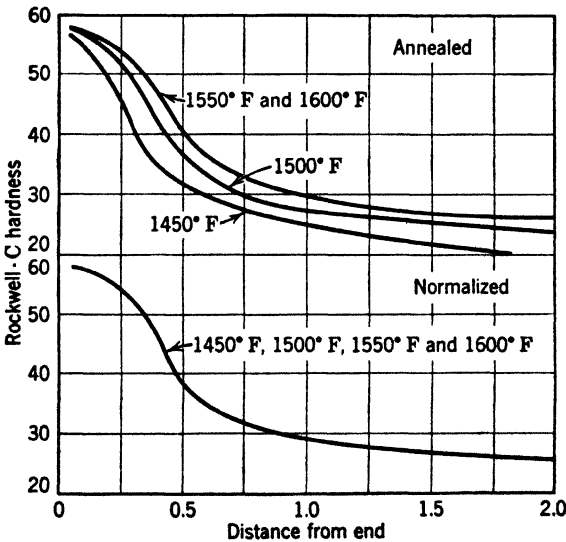


FIG. 196. Effect of quenching temperature on the hardenability of N.E. 9442 from two prior structures.

1600°, and no grain growth was shown. But the annealed 9442, with coarse pearlite and some of the carbide as spheroids, required 1550° to dissolve these larger carbide particles. Once they were dissolved, the Jominy curve was the same as with the initially normalized structure. Bain<sup>11</sup> illustrates this for eutectoid steel, Fig. 197. There was evidence that C diffusion was rapid but alloy-

ing-element diffusion slow, particularly noticeable in the highly alloyed 3240 and 4340.

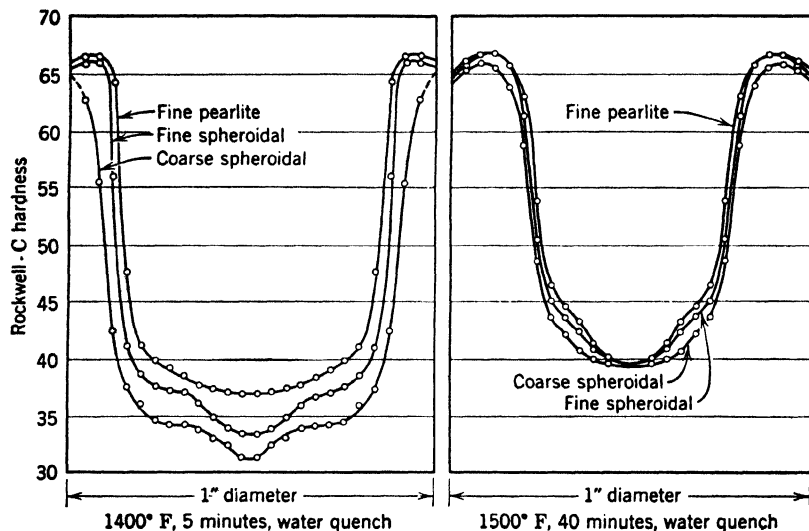


FIG. 197. Hardness distribution in 1-in. rounds of eutectoid steel as influenced by structure entering heating bath for constant heating schedule. Left—short time at 1400°F; right—longer time at 1500°F.

**Carbide Solution.** In ordinary heat treatment, appreciable time is given for homogenization, but in surface hardening, especially when the heating is done by electric induction, the time permitted is very short.

Using a triple-alloy steel of 0.63 C, 0.30 Si, 1.83 Mn, 0.75 Cr, 32 Mo, 0.80% Ni, Ellis<sup>12</sup> studied the results of very rapid austenitization by high-frequency heating, involving heating from the cold to 1575° in 130 sec and drastically quenching 30 sec later, after the temperature had dropped slightly. Starting with a fully spheroidized initial condition, not all of the carbide was dissolved in austenite, and yet the steel hardened to 62–63 RC. Starting with a pearlite structure and the same treatment, carbide was apparently dissolved, but either the C or some of the alloy did not completely diffuse, for, though the quenched hardness was 62½ RC, the quenched structure often contained striated formations closely resembling the initial pearlite structure. As Fig. 40 shows, martensite need contain only 0.45% C to obtain this hardness; moreover even a Rockwell-C indentation is large enough so that average hardness of higher- and

lower-C layers would be measured. The stratified structure, pseudo-morphic after or closely resembling that of original pearlite, could possibly be due to different etching of high- and low-C regions, but seems more probably to be chiefly due to lack of complete diffusion of the alloying elements.

**Induction Hardening.** That the hardness of quenched steel was about the same whether all the carbide had been dissolved or not is also explained by Fig. 40. Without definite proof, Ellis suggests that the heterogeneous martensite resulting from quenching the inhomogeneous austenite, has less internal stress than does ordinary acicular martensite; and suggests that the normalizing usually recommended as pretreatment for steel to be inductively hardened is a step in the wrong direction. This suggestion, however, supposes that a less hardenable steel more commonly used for surface hardening would not be adversely affected by incomplete solution and homogenization, as most observers would expect it to be. Poynter<sup>13</sup> presents experimental evidence for high internal stress rather than low stress. He heated 4340 by induction, transferred the specimens to an oil-quenching bath, the operation taking 3 sec (the S curve shows that a 10-sec delay is allowable). The steel coarsened just the same as when heated more slowly to the same temperature in ordinary furnace heating. Poynter suggests that measurements of surface temperature by welded-on thermocouples are inaccurate, through local heating due to the irregularity of contour at the couple, and so he measured temperature with an optical pyrometer. In such oil quenching the surface hardness was identical with that secured by oil quenching from regular furnace heating. But when the induction-hardened piece was water-quenched it did have hyperhardness, and, when such superhard specimens were cut up and stress thus relieved, the hardness of the surface near a cut notably decreased, evidence of the prior presence of internal stress. Martin and Van Note<sup>14</sup> consider that the lack of complete austenitization in high-frequency heating leads to easier transformation and thus avoids retained austenite. Steels of high hardenability are considered too likely to crack; 1350 and 2330 are favored by these authors.

**Homogenization.** Morris and McQuaid<sup>15</sup> say, "We can safely state that for maximum hardenability in a given [hypoeutectoid] analysis the C must not only be in complete solution but must be uniformly diffused." Bain<sup>11</sup> says, "For practical purposes hardenability is impaired by inhomogeneity resulting from insufficient heating," then goes on to advocate a prior normalizing treatment.

The apparent toughness of untempered heterogeneous martensite, noted by Ellis, is referred to by Focke, who, in discussion, remarks that 1045 steel is inductively hardened without the brittleness of an ordinarily quenched martensite, and speculates whether this may be due to heterogeneity of the martensite. Offhand, it would appear that the matter of compression rather than tensile stress in the martensite layer would be the thing to evaluate first, as indicated by Poynter's data.

Full carbide solution by austenite is necessary for maximum hardenability of the C and alloy steels most often quench-hardened, and normalizing is an easy method of producing an initial structure favorable to easy solution. Diffusion of C is easier than that of many of the alloying elements, and the necessity for a high degree of homogenization will doubtless vary with composition. Ham<sup>16</sup> finds that diffusion of Mo, for example, goes on three times as fast at 2000° as at 1800°. It is reported<sup>17</sup> that a surface layer of Cr, deposited on steel by chemical displacement from a fused salt bath, very readily takes up and diffuses C, but, if a little V is codeposited, the Cr layer no longer shows rapid C diffusion. The diffusion conditions in alloy steel may thus be quite complex.

**Hypereutectoid Steels.** Hypereutectoid steels, such as ordinary plain-C tool steel, contain excess cementite over the pearlite ratio. One purpose of an anneal in this grade of steel is to secure machinability. The as-rolled steel is rather hard, say, 225 Brinell, and contains a cementite network surrounding coarse laminated pearlite grains. With this structure and hardness most types of machining are not so easily done as with the steel in softer condition and with its cementite more or less spheroidized.

As Fig. 3 shows, the  $A_{cm}$  line rises steeply, and to take the steel sufficiently above it to bring all cementite into solution and fully homogenize the austenite might lead to unhealthy grain growth. Moreover, the presence of well-distributed tiny cementite globules in the quenched and tempered matrix adds to the cutting power of the tool through improved wear resistance. Hence, full solution of cementite is not always necessary, and austenitization is ordinarily carried out somewhat below the  $A_{cm}$  line in the higher-C steels; whereas, in hypoeutectoid steels it is very necessary to go above  $A_{c_2}$ .

However, behavior on cooling is greatly modified by full instead of partial solution of the carbide. This is especially true in regard to hardenability secured on quenching, but it also applies to annealing.

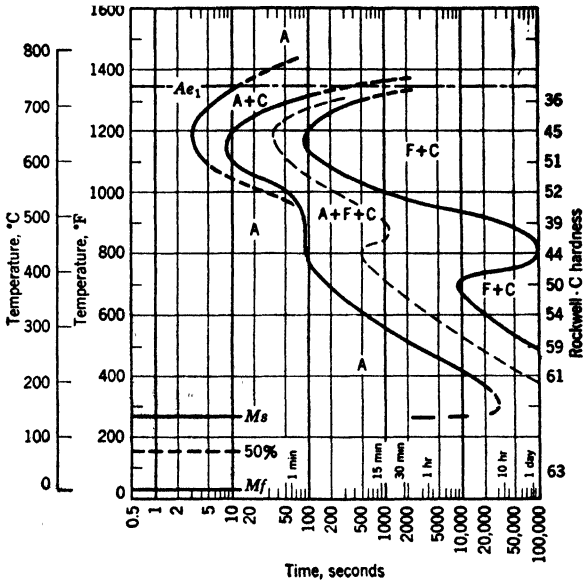


FIG. 198. Carbides dissolved.

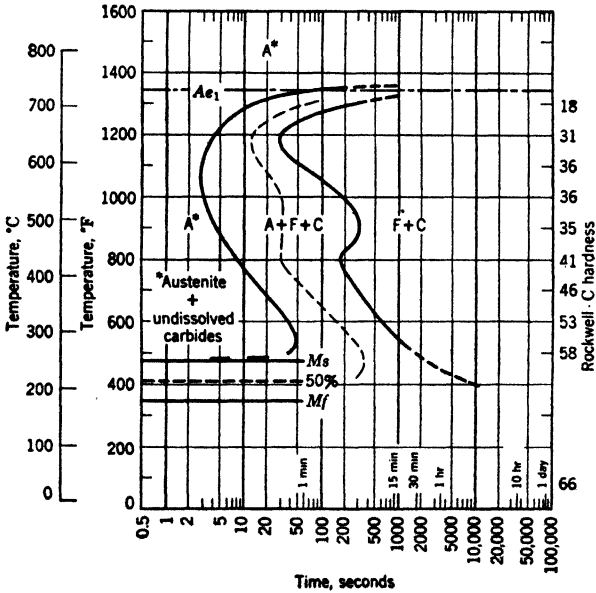


FIG. 199. Carbides not dissolved.

SAE 52100, containing 1.02 C, 0.36 Mn, 0.33 Si, 0.20 Ni, 1.41% Cr, austenitized at 1550°F; grain size: 9. (U. S. Steel Corp.'s *Atlas of Isothermal Transformation Diagrams*)



As Figs. 198–199 show (right-hand margin), keeping the austenite from picking up its full quota of C makes the resultant annealed structure softer. The fully homogenized 52100 takes 100 sec instead of 30 sec to complete the transformation at 1200° and shows a hardness of Rockwell C-45 against 31. The undissolved carbide makes the reaction go on faster, by the “trigger action.” The difference in hardness after holding at 1300° is noteworthy, Rockwell C-36 for dissolved, and only 18 in the case of undissolved carbides.

The presence of carbide particles leads to the separation on cooling of more carbide upon these particles, increasing their size instead of its coming out in an intergranular network or in spines. The  $M_s$ – $M_f$  range is greatly altered. Analogous behavior for another steel is shown in Figs. 201–202.

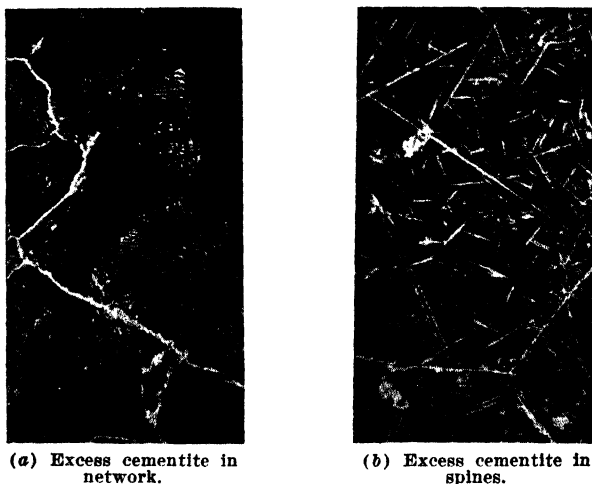


FIG. 200. Examples of poor initial structure in high-C tool steel. (Bullens)

**Cementite Envelopes.** As is discussed under spheroidization (pp. 54, 414), a preliminary austenitization above  $A_{cm}$  is called for when the steel has cementite envelopes, or spines extending into them, about the pearlite grains, owing to high finishing temperatures or insufficient working, such as the structure of Fig. 200. Once the cementite is taken into solution, holding between  $A_{cm}$  and  $A_{c1}$  will precipitate some carbide. Further cooling and holding at, or reheating to, a temperature just under  $A_{c1}$  produces further spheroidization and softening. Spheroidization of hypoeutectoid steels is facilitated by austenitizing below the end of the critical range on

heating, so that some untransformed ferrite remains; this is just the opposite of the proper pretreatment for hardening.

**Utility of Undissolved Carbide.** In normalizing to produce a fine and hence a strong structure, fine structure is obtained by shoving the pearlite nose of the S curve down to a lower temperature and back to a longer time delay. This is favored by homogenizing the austenite. But, once the austenite has been preserved as such down to the temperature corresponding to the structure desired, one wants the transformation to take place rapidly and as nearly as possible, concurrently throughout the whole section. When it is time to pull the trigger, trigger action is needed to carry the transformation rapidly through a grain or through the section, but low-C austenite boundaries do not provide the kind of trigger action needed, since they will transform at too high a temperature.

Hence we need austenite well homogenized as to C, but still fine, without grain growth, and carrying some other trigger. This can well be a small amount of undissolved carbide left when the austenitizing time and temperature are just short of full solution. Vanadium is often used to provide such a carbide. Additions of V of the order of 0.05%–0.10% are sufficient. Titanium, and, though probably to a lesser degree, zirconium, or other strong carbide formers, behave quite similarly, and in as small amounts. The possibility is not entirely excluded that some stable oxide, nitride, or other compound, rather than carbide, is responsible, nor is it certain whether V and Ti enter into cementite, making it more stable and less soluble, or exist as separate carbides. The general behavior is much as though the responsible factor were a separate carbide.

It appears as though everyday cementite,  $(\text{FeMn})_3\text{C}$ , were rather easily dissolved by austenite and its C rather readily diffused to produce a relatively homogeneous austenite, which thereby has its  $A_{r_{3-2-1}}$  depressed, but the vanadium or titanium carbide is far less soluble. After austenitization sufficing to dissolve  $(\text{FeMn})_3\text{C}$ , the vanadium or titanium carbide remains, and its presence inhibits grain growth. That the small amounts found efficacious could be expected to be so, is indicated by the equally small amount of Al (generally supposed to be present as submicroscopic  $\text{Al}_2\text{O}_3$  and  $\text{AlN}$ ) required for grain-size control.

When austenite containing what is assumed to be undissolved vanadium or titanium carbide is more vigorously austenitized, by heating to a still higher temperature, grain size increases rather rapidly, quench hardenability increases, as is shown by the posi-

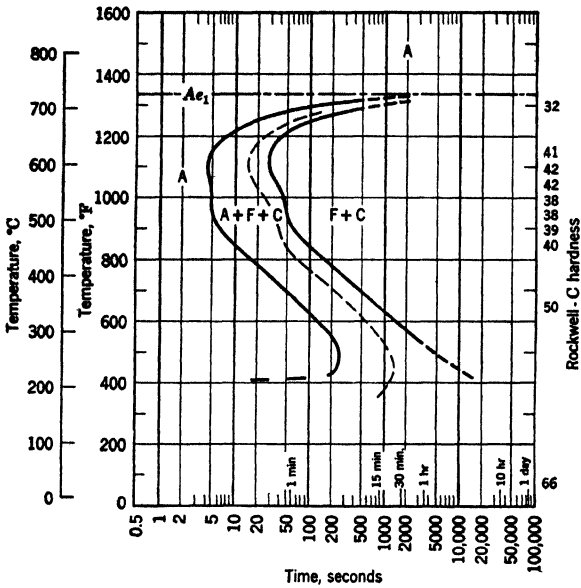


FIG. 201. Carbides dissolved.

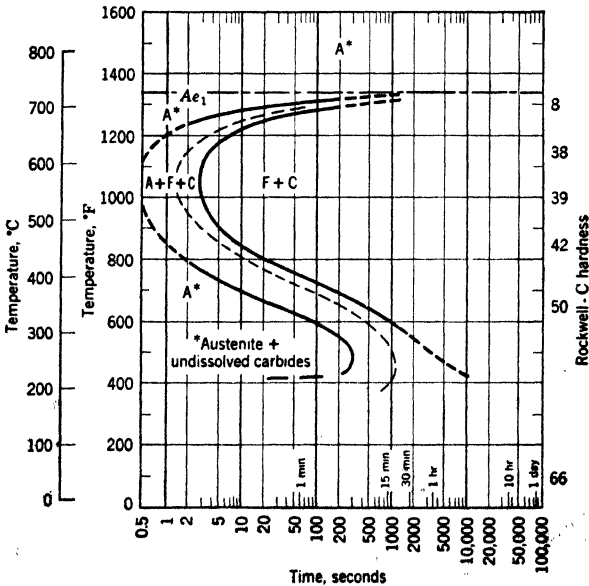


FIG. 202. Carbides not dissolved.

0.3% V steel, containing 0.87 C, 0.30 Mn, 0.27% V, austenitized at 1500°F; grain size: 11 (U. S. Steel Corp.'s *Atlas of Isothermal Transformation Diagrams*)

tion of the pearlite nose in Figs. 201–202, and, on tempering a quench-hardened piece, a secondary hardening, of the precipitation-hardening type occurs, all strong indications that the vanadium or titanium carbide has now been taken into solution. The temperature at which rapid grain growth occurs is somewhat lower in the case of V than in that of Ti, and this is in turn still lower than in the case of Al used as a controller of grain growth. However, the temperature in the case of V is sufficiently above that for solution of  $(\text{FeMn})_3\text{C}$ , so that heating for quenching or for normalizing, done at the usual temperature for such a steel from which V is absent, leaves the vanadium carbide at least partly undissolved; but, when coarsening is desired, a moderate increase in temperature accomplishes it. Vanadium thus allows easy control, and, as a grain-growth retarder is somewhat more fool-proof than aluminum, since there is always ample carbon to combine with vanadium; whereas, in the use of aluminum, the oxygen, sulphur, and nitrogen levels must be adjusted before the aluminum is added, in order to get the desired effect.

The untangling of homogenization from the accompaniment of grain growth was a major step in the development of pre-heat treatment. Next we turn to the matter of grain size and grain growth.

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## CHAPTER 15

### AUSTENITIC GRAIN SIZE AND GRAIN GROWTH

Observations on the tendency of austenite to grow large grains by overheating, on the effects of overheating on the final product and on the individual peculiarities of different heats of the "same" steel were made early in the history of heat treatment. Not until the precipitation hardening of metals began to be understood was any logical explanation adduced to tie these various behaviors together.

The metallurgical pioneer, Henry M. Howe, clearly showed the effect of increasing temperature in the austenite range, in Fig. 203, for normalizing, and Fig. 204, for annealing.

The general effect of coarse grain in decreasing the notched-bar toughness of steel has been indicated in Fig. 110. Such knowledge led to early warnings, emphatically stated in the first edition of this book, against overheating of austenite, as exemplified in Fig. 144. At that time the general run of steels was easily coarsenable, and so the warning was made general. Later it developed that steels and individual heats of steels varied greatly in propensity toward coarsening, and so the warning was modified in regard to those that were difficultly coarsenable.

In very early days it was recognized that some steels, especially high-C tool steels, varied in hardenability, even though their chemical analyses did not vary. The ability to harden readily was denoted by the term "body," or "timbre."

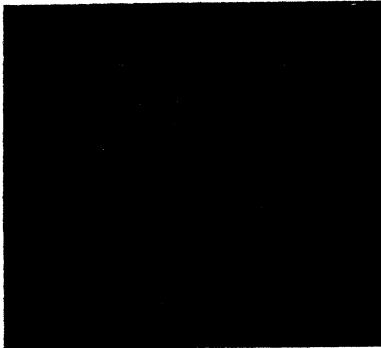
This variation in hardenability was known for many types of steels, but experimental attention was focused on it in relation to carburized steels. In quenching the high-C surfaces of box-carburized steels in oil or in warm water, it was found that some steels were prone to show soft spots, instead of the expected fully hard martensite all over. This means that the critical cooling rate for full hardening was greater than was locally reached. The cooling rate attainable was just on the ragged edge of that required, and a film of scale or a steam bubble adhering to the steel would throw



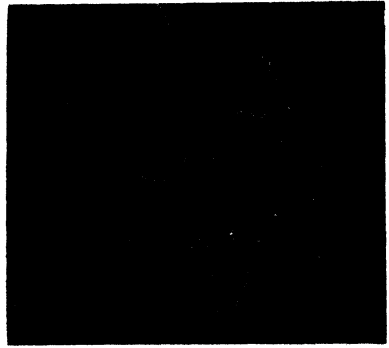
1470°F



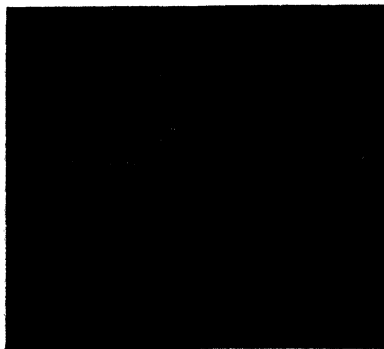
1650°F



1830°F



2010°F



2190°F

X40. (Howe)

Fig. 203. Effect of heating beyond  $A_{c3}$ . 0.40% C steel heated at temperatures indicated for 10 min and air-cooled.

the cooling rate of the steel beneath it, below that for martensitic hardening. More drastic quenching was required on such steels to avoid soft spots.

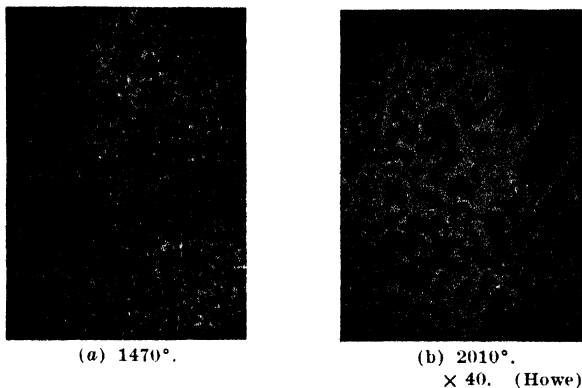


Fig. 204. Effect of heating a 0.40% C steel for 10 min at temperatures shown, followed by furnace cooling.

**Normal and Abnormal Steel.** The issue was somewhat confused by another phenomenon. The steels that gave soft spots had the cementite of the hypereutectoid zone coalesced into patches instead of the grains being outlined as a network in the normal fashion, when the steel was cooled slowly from the carburizing temperature. The type with the network was called "normal"; the other type, "abnormal" steel. At that time, because of the prevalence of soft spots in carburized "abnormal" steel, the implication that "abnormality" connoted inferiority was not disturbing, and the nomenclature seemed natural. Further experience showed how to avoid the troubles with the "abnormal" steels and showed, moreover, that they had many virtues, so that today, although each type is superior in certain fields, the "abnormal" steels would probably be classed as so useful a product that some name implying superiority rather than inferiority would be chosen, were they to be renamed. Moreover the method of carburizing may affect the "normality," so that one should use the terms to refer to a structure rather than to a steel. This was brought out in Fig. 60. by the difference between gas- and pack-carburizing.

Cases where appraisal of normality has been invoked in connection with behavior of steel are given by Wolfe,<sup>1</sup> who finds that coarse grain and abnormal structure accompany poor deep-drawing



properties in autobody sheet, and by Kerr and Eberle,<sup>2</sup> who find the normal structure more resistant to graphitization.

The early studies were chiefly on carburized steels. In these, on sufficiently slow cooling from the carburizing temperature, the hyper-eutectoid cementite was found to outline what had been, at the carburizing temperature, very large austenite grains. Such steels were invariably deep hardening. Steels with fine grain (and, as later found out, whether their cementite tended to coalesce or not) were found to be shallow hardening. To evaluate hardenability, and to differentiate between normal and abnormal steels, the McQuaid-Ehn test was introduced. In time the idea of the effect of austenitic grain size became untangled from that of the distribution of the cementite, that is, from the "normality," and grain size of austenite began to be studied as a major variable.

**Experience with Vanadium Steels.** At this early date grain size just happened, it was not intentionally controlled in ordinary C steels. However, there was a large background of experience with V steels, in which quite small amounts of V, 0.05 to 0.15%, were known to produce very tough fine-grained steels, this being especially noticeable in normalized steels where, in the absence of V, a high normalizing temperature was conducive to grain growth. Other alloying elements, particularly Mo and Ni, were known to have some inhibiting action on the grain growth of austenite.

**Aluminum Additions.** In those days final deoxidation of steel with a trace of Al, say an addition of 0.01 to 0.05%, was often employed to quiet a "wild" heat of steel. Need for Al was considered a sign of sloppy steel making and the steel makers who did use it were very loath to admit such practice (compare McQuaid<sup>3</sup>). The melters would even provide Al at their own expense and use it to get sound steel against the instructions of the management. When Al was used, the steel had a decided tendency to be fine grained and also abnormal. Epstein and Rawdon<sup>4</sup> clarified the subject considerably. They subjected a series of steels, varying in "normality" by the McQuaid-Ehn test, to spectroscopic analysis for Al (which would not differentiate between Al present as the metal and as oxide or nitride) and found that fine grain and abnormality were connected with the Al content. An Al-free steel was coarse grained, normal, and deep hardening, whereas the Al-treated steels were the reverse. Epstein then added V and Al to a couple of ingots in a steel mill and compared these ingots with untreated ingots from the same heat. He found that Al acted very much like V in producing fine grain,

and pointed out that steels having much the same properties as V steels could be made at a lower cost, with controlled additions of Al. Commercial practice now utilizes Al very widely. A difference between the action of the two elements is that the V-treated steels are not prone to show the "abnormal" coalescence of cementite when carburized.

Later other strong deoxidizers such as Ti and Zr were appreciated as likewise capable of producing fine-grained austenite.

**The McQuaid-Ehn Test.** We have spoken as though the austenite of the fine-grained steels was necessarily fine grained at any temperature in the austenite field and such an idea is more or less implied in much of the published discussion; but this is quite incorrect. The idea has sprung from the widespread use of the McQuaid-Ehn test\* at one temperature, 1700°, only. Since most carburizing is carried out at 1700° or a little below, the test is adequate to show how the steels being examined will act in carburizing. When the McQuaid-Ehn test is used to evaluate the behavior of a carburizing steel, the carburizing temperature, time, and medium that are to be used in practice ought to be adhered to. Houdremont and Schrader,<sup>5</sup> among others, have shown that on carburizing for the same time and temperature in a solid carburizer (BaCO<sub>3</sub> and charcoal) and in coal gas, several members of a series of steels studied behaved quite differently in regard to grain growth. The nature of the carburizing medium, therefore, cannot be disregarded.

**Inadequacy of the McQuaid-Ehn Test.** When the problem is not one of carburizing but of evaluating the steel as to coarsening propensities for other uses, the McQuaid-Ehn test cannot be considered as necessarily adequate, for, as Fig. 208 shows,<sup>6</sup> failure to coarsen at 1700° by no means ensures that the steel will not coarsen at a higher temperature, such, for example, as would be used in patenting. Nor does it tell whether a steel that does coarsen in the McQuaid-Ehn test might not have failed to coarsen at some lower temperature such as would properly be used in heating for quenching. Notwithstanding this obvious limitation and the fact that it would be more logical to apply the test at the exact temperature at which the steel is to be heated, if that is known, or over a range of temperatures if the object is to select the proper temperature, the "inherent austenitic grain size" used often to be allegedly described in terms of the grain size produced at 1700° only.

\* Box carburizing with solid compound of the type that will produce a hypereutectoid case, for 8 hr (usually at 1700°), followed by slow cooling.

**Standard Grain Size.** The American Society for Testing Materials has set up a series of eight grain sizes as seen under the microscope by which the steels are designated.\* The series may be extended at each end. Sizes of 0 and -1 on the coarse end and 9 and 10 on the fine end sometimes are reported. Figure 58 showed the size of these grains at  $\times 100$ . In less precise terms, sizes 1 to 5 are classed as coarse and sizes 6 to 8 as fine. The grain size of the ASTM series varies in a regular progression, the number of grains per square inch on a section examined at  $\times 100$  doubling at each grain number. As Herty, McBride, and Hollenback<sup>7</sup> point out, it is the grain surface in three dimensions that is the real factor, and this varies as shown in Table 27.

TABLE 27

ASTM Grain No.	No. Grains per Sq In. on Section Viewed at $\times 100$	Grain * Surface	Inter- <sup>†</sup> facial Area
1	1	4.0	270
2	2	5.6	340
3	4	8.0	480
4	8	11.3	679
5	16	16.0	961
6	32	22.6	1360
7	64	32.0	1920
8	128	45.3	2720

\* Square inch of grain surface per cubic inch at  $\times 100$ , calculated for cubical grains. Herty, McBride and Hollenback.

<sup>†</sup> Mean total interfacial area square inch per cubic inch calculated for a 14-sided solid of maximum ability for close packing. From Rutherford, Aborn, and Bain.

What the grain size really implies, and the factor that counts most, is the grain-boundary area. Johnson<sup>8</sup> suggests a refinement in grain-size determination, aimed to bring out the average and the spread in the three-dimensional grain size, along the principles set forth by Herty and coworkers.

The details of Johnson's methods of measurement and computation are too complex to repeat here. When the data are at hand,

\* The official grain size chart of the American Society for Testing Materials showing hypereutectoid and hypoeutectoid zones of the case of steel subjected to the McQuaid-Ehn box-carburizing test with solid carburizer and slow-cooled to develop the cementite and ferrite networks, respectively, is given in ASTM Standard E-19-33. This specification may be purchased from the American Society for Testing Materials, 1916 Race Street, Philadelphia 3, Pa.

they can be plotted, in probability fashion, to show the size distribution, also the grain-boundary area.

**Three-Dimensional Grain Size.** Since the grain count or estimate is made on a plane section viewed at  $\times 100$ , it is obvious that this plane will cut the mass of crystals so that some crystals will show their maximum cross section whereas others will show only a small cross section where the plane cuts across a corner of the crystal. Hence "mixed grain," as Rutherford, Aborn, and Bain<sup>9</sup> show, is not evidenced unless the smaller grains are grouped in clusters. When the cross section merely shows smaller grains rather regularly interspersed among the larger grains, the effect is due to the geometry of the section. When true mixed grain appears, it is an indication either of inhomogeneity of the steel or that the coarsening process was caught in an active state, coarsening having begun but not proceeded to equilibrium for the temperature used, so that the absorption of small crystals by larger ones was irregular. Figure 207 shows mixed grain.

**Coarsening Characteristics.** It is thoroughly illogical to try to evaluate all the grain size and grain-growth characteristics of a steel by a test under one condition only, as by applying the McQuaid-Ehn test at 1700° alone. A figure that tells more is the temperature range of coarsening, but that figure likewise fails to tell the whole story. More complete data, including the austenitic grain size at each of a range of temperatures below and above the temperature at which notable coarsening begins, are necessary to describe the "coarsening characteristics." The situation is comparable to endurance testing. One cannot make any broad comparison of steels on the basis of life at one load; a series of tests at different loads is necessary to establish the S-N curve.

The term "inherent austenitic grain size" is a misnomer, for no single grain size is inherent. It may occur that two steels treated the same way and having, under that treatment, the same grain size will behave differently and in ways that seem to be related to the fact that one of the steels would have coarsened and the other would not, had they been heated to a higher temperature. That is, the nature of the steels, which is evidenced by variations in ability to coarsen, also produces other variations in behavior even though the coarsening tendency has not yet come into play.

**Opportunities for Rejection at Boundaries.** Since, in hypoeutectoid steels, ferrite, and in hypereutectoid, cementite is rejected at the grain boundaries of austenite as it passes through the critical

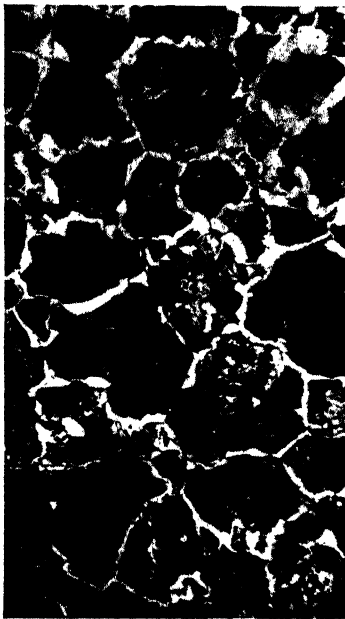
range, a pure grain-size effect is only present in eutectoid steels. In others the lining of the grain boundaries with a different material has its effect. Herty and coworkers<sup>7,10</sup> point out that the rate of cooling through the critical affects the amount of boundary ferrite or cementite. With hypoeutectoid steels the amount of ferrite correlates closely with the impact resistance. So, although it is ordinarily phrased that coarse-grained and fine-grained steels behave thus and so, this means only that the cited behavior holds for the steels cooled at the certain rate being discussed.

**Outlining the Grains.** The austenitic grain size is fixed by the heating. It may or may not be clearly observable in the steel as prepared for service; but, if it is not, some method that will show that grain size must be adopted. The rate of cooling will alter the way the austenite grains are outlined, by cementite, ferrite, or pearlite, according to the composition of the steel. Regardless of the way the steel is to be cooled for its regular heat treatment for service, the grain size established in its heating is fixed; various legitimate ways may be chosen to make this size apparent. In the McQuaid-Ehn test the composition of the steel is altered by introduced C, and the time required for C diffusion may be far longer than the heating time for any treatment except carburizing.

**Other Methods for Delineation of Grain Size.** Slow cooling from the correct heating time and temperature will allow cementite of hypereutectoid (see Fig. 60) and ferrite or pearlite of hypoeutectoid (see Fig. 205) steels to separate and outline the grains; but often it is better, with a hypoeutectoid steel, to cool down into, but not past, the critical range and then quench. This is termed an "arrested quench." (See Fig. 206.) At compositions near the eutectoid some other means must be found. Some conditions of quenching and tempering can usually be found that will serve.<sup>7</sup> If one end of the specimen is water-quenched and the other left out of the water, the zone between the hardened and unhardened parts will have pearlite nodules outlining the grains. This is termed a "differential quench" (see Figs. 206, 207). Light tempering of fully hardened steel to "secondary troostite" may also be used. In any case, a suitable etching reagent must be found that will show contrast between grains and boundaries.

Quenching from the temperature in which one is interested, tempering lightly, and etching to develop the martensitic grain size that reflects the austenitic grain size, is described by Bain and Välella,<sup>11</sup> and is much to be preferred to a test at an arbitrary temperature.

Heat tinting is sometimes employed to bring out the grain boundaries. Eutectoid steels and some alloy steels may not readily etch or heat-tint in such fashion as to show the austenitic grain size clearly, so that it is occasionally necessary to work out a combination method. Tobin and Kenyon<sup>12</sup> carry out the heating under slightly oxidizing conditions so as to attack the grain-boundary ma-



0.47% C.  
Coarse, 1800°.



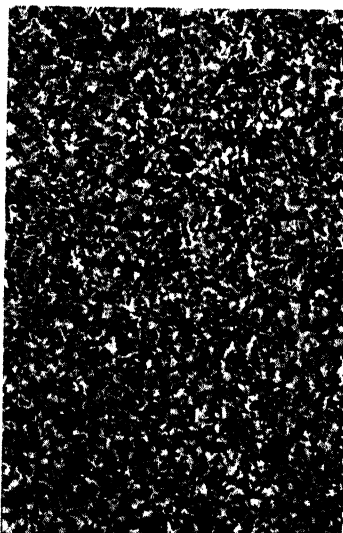
0.85% C.  
Fine and duplex, 1700°.  
× 100. (Lauderdale, Battelle)

FIG. 205. Cores of specimens box-carburized 8 hr at temperatures shown, and slowly cooled. The grains are outlined by ferrite.

terial and render it more strongly attacked in the etching process. Day and coworkers<sup>13</sup> use heat etching in H, with special precautions. Their method appears applicable to all steels. Compare Hawkes.<sup>53</sup>

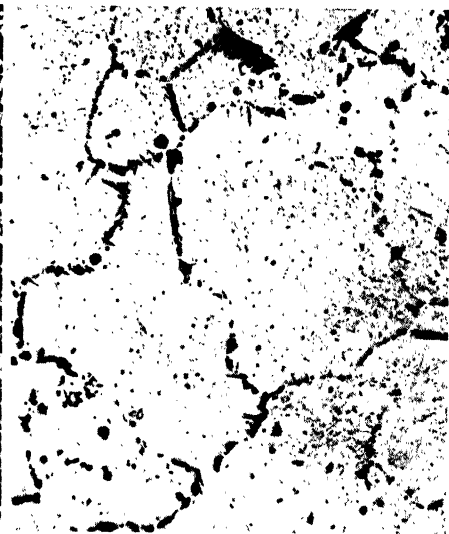
**Grain Size in Tool Steels.** High-C tool steels, quenched and fractured without tempering, demonstrate their grain size to visual examination.

Shepherd and the Swedish Jernkontoret both developed sets of tool steel standards, varying in austenitic grain size from 0 to 10 on the ASTM scale. By comparison with these standards,<sup>14</sup> the grain size of a C or low-alloy tool steel can be visually noted without resorting to the microscope. The "fracture grain-size" numbers, as-



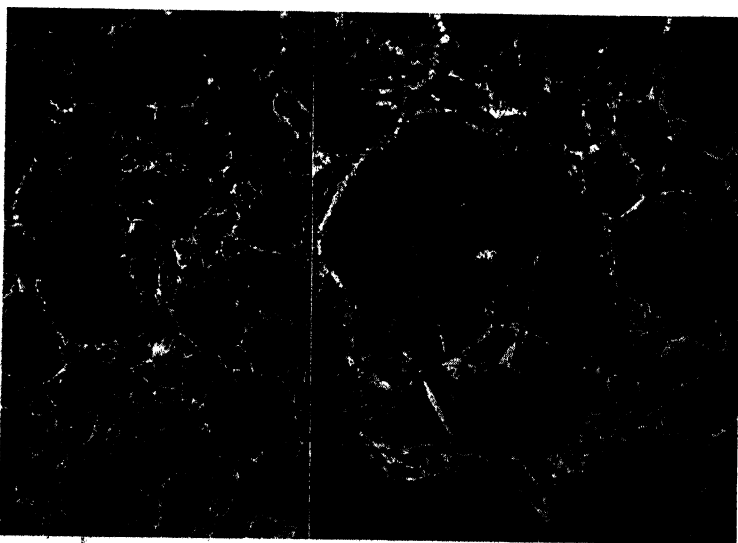
Fine grain.

0.45% C steel. Heated 1 hr at 1650°. Differential quench.



Coarse grain.

0.45% C steel. Heated 8 hr at 1920°. Arrested quench at 1380°.



Medium grain.

0.41% C steel. Heated 1 hr at 1650°. Differential quench.

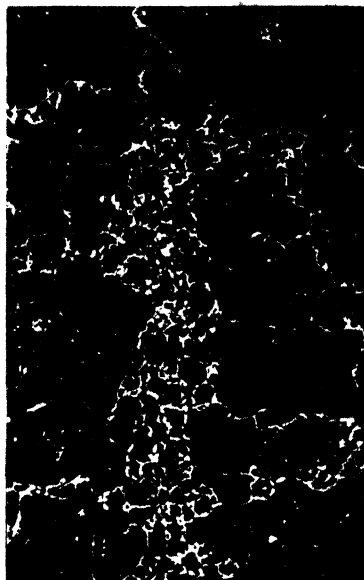
Coarse grain.

0.41% C steel. Heated 1 hr at 1750°. Differential quench.

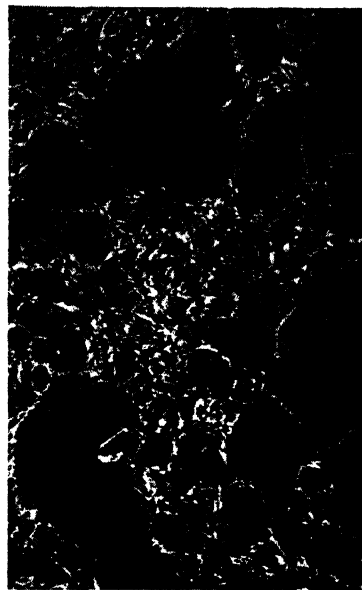
All  $\times 100$ . (Lauderdale, Battelle)

Fig. 206. Grain sizes delineated by quenching methods.

signed to the standard comparison specimens, are numbered like the ASTM series, and the numbers agree closely. The fracture grain size can be read only on the fractures of martensitic steels of the tool type and would be in error on low- or medium-C steels. However, Enlund and Schlberg<sup>52</sup> have discussed the application of the fracture method to lower-C steels. They note that subtracting the Jernkontoret number from 16 gives the ASTM number.



1.10% C tool steel, held 8 hr at 1700°. Slow cooled, nital etch.



0.41% C steel, held 1 hr at 1650°. Differential quench.  $\times 100$ . (Lauderdale, Battelle)

Fig. 207. Duplex, or mixed, grain size.

Grossmann<sup>15</sup> points out that the Shepherd fracture-test series of specimens are C tool steels and give useful information on the C tool steels to be examined, but may give quite erroneous indications of uses for comparison of fracture-grain sizes in other steels, a detailed example being given for 8640. Quenching from a range of temperatures and determining the fracture grain size evaluates the grain-coarsening characteristics of tool steels.

In the case of high-speed tool steel, which is usually very fine grained, Snyder and Graf<sup>43</sup> suggest examination at  $\times 1000$  and counting of the grains intercepted by a diameter of the field, after etching. Such steels, in a condition for service, may have the aus-



tenitic grain size revealed by etching with up to 10% HCl and 3% HNO<sub>3</sub> in methyl alcohol. This method allows discrimination among steels whose grain sizes correspond to from 8 to 10 on the extended ASTM scale, which, if compared visually at  $\times 100$  with the usual chart, could only be classed as "finer than grain 8." The cutting properties are influenced by variations in this range of grain size.

**Hardenability of Tool Steels in Respect to Grain Size.** Tool steels have long been considered to show such mysterious variations in behavior of different lots that purchase by brand name was the rule. However, the mysterious "body" of C tool steels has been stripped of most of its mystery by the application of knowledge of grain-size effects. For example, Digges and Jordan<sup>44</sup> studied two 1% C steels, which were essentially identical in chemical analysis for all elements usually determined, that is, differing chemically only in that one contained 0.01% Al instead of a scarcely detectable trace, but which showed entirely different depth-hardening properties. They found the differences thoroughly explained by the grain size, grain-coarsening characteristics, and resultant differences in critical cooling rates.

Schempp and Shapiro,<sup>45</sup> working with 1.05% C steels, found that cold working, and to some degree, prior heat treatment, alter the grain-coarsening characteristics. In the discussion of this paper it was clearly brought out that other factors, such as those mentioned by Schempp and Shapiro, may be superimposed on the "inherent" tendencies of the melt itself, to modify the behavior of the finished steel. Portevin<sup>46</sup> has discussed the effect of cold work.

Palmer<sup>16</sup> brings out the fact that in high-C tool steels, depth of hardening and propensity toward grain growth do not necessarily go hand in hand. Moreover, the fracture of a hardened specimen quenched from a normal temperature, say 1450°, may not reveal differences in two tool steels (they may both appear fine grained and show equal depth of hardening), and so he advocates quenching in brine from 1550° to reveal the differences. This increase in quenching temperature is more likely to bring out the innate differences. The one that shows coarsening from overheating in such a test will be inferior in toughness even when quenched and drawn at the usual temperatures, the variation in toughness often being of the order of 100% when evaluated by the torsion impact test, while the hardness of the two steels is identical. Differences are also apparent in susceptibility to grinding checks. The more readily coars-

ened and more brittle steel will have tiny surface cracks when abused in grinding; the tougher one will stand more abuse.

**Coarsening Propensities as Vital as Actual Grain Size.** Boulger\* has also found instances in medium-C steels where the impact re-

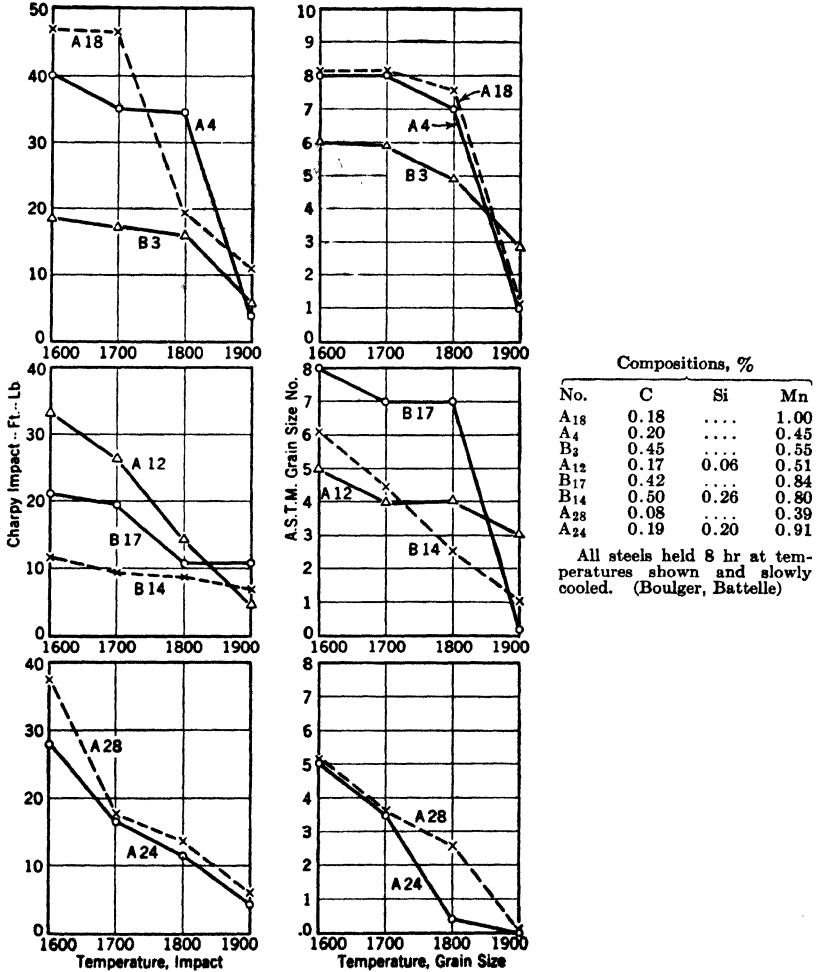


FIG. 208. Comparison of impact and grain size of annealed steels.

sistance of lower-C steels falls off at a temperature approaching the coarsening temperature but still too low to show more than a very little coarsening. Thus in Fig. 208, from his work, the shape of the impact and grain-size curves is similar, taking into account the

\* Unpublished work, Battelle Memorial Institute.

initial differences in impact due to chemical composition; but in the samples of A12, A18, and B17 annealed at 1800° a drop in impact occurs that is disproportionately great in relation to the degree of coarsening at 1800°.

**Grain Size and Hardenability.** The outstanding importance of grain size in heat treatment rests in the relationship to depth of hardening on quenching. The relative hardenabilities resulting from differences in grain size were shown in Fig. 173, from the early work of Grossmann.<sup>17</sup> The effects are shown in terms of ideal diameter, on the 50% martensite criterion, but the relative behavior indicated appears to hold, whatever the criteria. What is really evaluated is the relative time delay at the pearlite nose of the S curve for continuous cooling. Hollomon and Jaffe<sup>18</sup> consider that grain size has a less notable effect on avoiding the production of bainite, but this distinction has not been generally made.

The combined effect of differences in grain size and presence of undissolved carbide has been shown in Figs. 198-199 and 201-202.

**Associated Factors.** It is difficult to untangle the individual effects of various factors associated with grain size.

The degree of sluggishness conferred on the steel in austenitizing it depends on temperature (and time). Ordinarily the steel is quenched direct from the austenitizing temperature; the phrase "quenching temperature" usually implies that the steel was also austenitized at that same temperature. However, the rate of cooling through the vital 1100-900° region of the pearlite "nose" must be affected by the temperature of the steel when it meets the coolant. Contradictory statements are met in the literature; one group argues that, starting from the higher temperature, the "temperature head" is greater; hence the *rate* of temperature loss at the surface in the vital pearlite-nose region will be more rapid; the other group says that it is obvious that cooling from, say, 1400 to 1000° is faster (takes less time) than from, say, 1700 to 1000°. Both appear to be right. But what we are most concerned with in considering depth of hardening is the rate of cooling in the 1100 to 900° range, *beneath* the surface.

Jackson and Christenson<sup>19</sup> point out that, *if* the grain size is the same at both quenching temperatures, the higher the temperature, the *less* severe the quench. This is much as though the hotter steel heated up the coolant in contact with it, which is accepted as decreasing the severity of quench. They showed this by austenitiz-

ing two Jominy bars at 2100°, quenching one direct from 2100° and the other (coarsened by the 2100° treatment) after cooling it to 1600° and holding it at 1600° for temperature equalization. On a shallow-hardening coarse-grained steel, of 0.30 C, 0.94 Mn, and 0.10 Cr, they obtained the Jominy curves of Fig. 171. The effect of decarburization on the surface skin, due to the high temperature, is

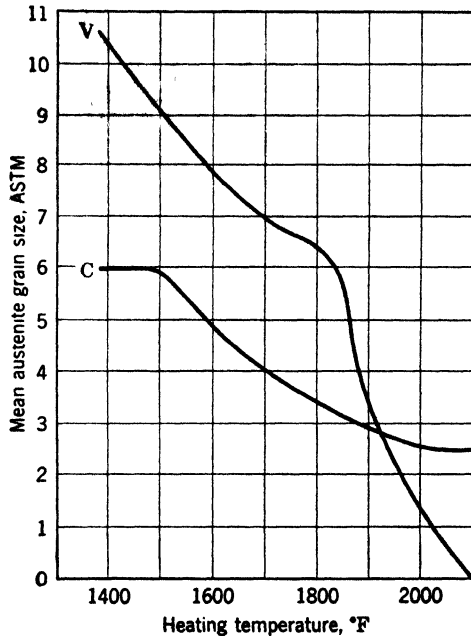


FIG. 209. The approximate trends of mean grain size in the austenite of two 0.90% C steels alike except for 0.27% V in steel V. Note coarsening ranges. (From data of Zimmerman, Aborn and Bain)

evident in these curves; also the maximum hardness is a bit low for the carbon content, and on three other coarse-grained shallow-hardening steels the curves are not consistent for the first  $\frac{1}{16}$  in. However, all four sets of curves are consistent from the  $\frac{1}{16}$ -in. position on, in that the 1600° quench curves lie above these for the 2100° quench.

Hence, we may conclude that, other things being equal, the higher the quenching temperature, the less the intensity of the quench, and that the usually observed fact, that the higher the quench temperature, the deeper the hardening, is due to the overbalancing effect of increased homogeneity and increased grain size.

A summation of the three effects, quenching temperature, homogeneity, and grain size (for medium-C and low-alloy steels they studied), is given by Clark and Richards<sup>20</sup> in Fig. 172, where increasing the austenitizing-quenching temperature increased the distance for Jominy-bar distance (at 50% martensite) as shown.

Increased time and temperature in the austenite field increasingly take carbide into solution, increasingly lead to greater homogenization, and increasingly tend to produce larger austenitic grain size. In "grain-size-controlled" steels, such as those to which adequate but fairly small additions of suitable elements have been made, this tendency is pretty much obliterated until a relatively high "coarsening temperature" is reached, when grain growth starts to occur at a rapid rate. Steels containing difficultly soluble carbides, such as vanadium carbide, tend to be fine grained at low austenitizing temperatures, the grain size slowly increasing with temperature (and time) as the carbide goes into solution, then increasing at a more rapid rate, as shown by the break at 1850° in the curve of Fig. 209. The degree of homogenization of austenite is simultaneously increasing, and so it is difficult to untangle the two factors. The grain-growth behavior of a few steels is shown in Fig. 210 from Dube and Gerstman,<sup>51</sup> which indicates that individual steels vary. Digges and Rosenberg,<sup>21</sup> Dorn and Harder,<sup>22</sup> and others have shown that the pretreatment, that is, the structure prior to austenitizing, has a marked effect on grain growth. The finer the preliminary carbide distribution, the more readily homogenized is the austenite, and the easier grain growth starts.

The effect of grain size on hardenability is indicated by Fig. 211.

TABLE 28

	C	Mn
	Per Cent	Per Cent
1	0.40	0.80
2	0.42	0.90
3	0.75	0.30
4	0.90	0.30

Aborn<sup>23</sup> shows Fig. 212 to indicate the effect of grain size on the time delay at the pearlite nose of the S curve. An apparent reversal of this behavior, noted by Erb,<sup>24</sup> for hypereutectoid tool steels, has been discussed on p. 303.

The grain-size effect on hardenability and on mechanical properties has been nicely shown by Schane,<sup>6</sup> Figs. 213-217. Two forg-

ing grade 1040 steels, similar in composition but differing in the way the heats were finished in the steel mill, were put through their paces.

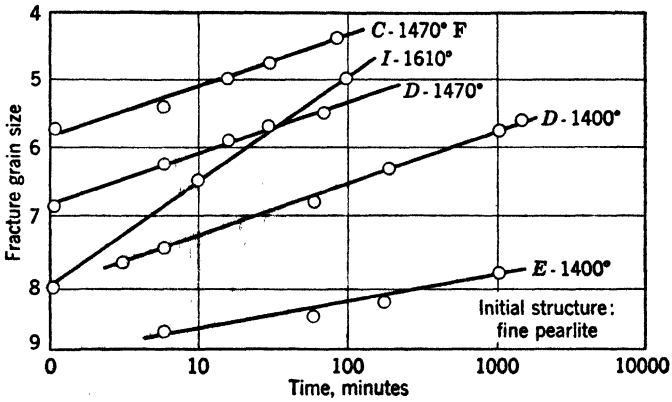


FIG. 210. Grain-growth behavior in different steels.

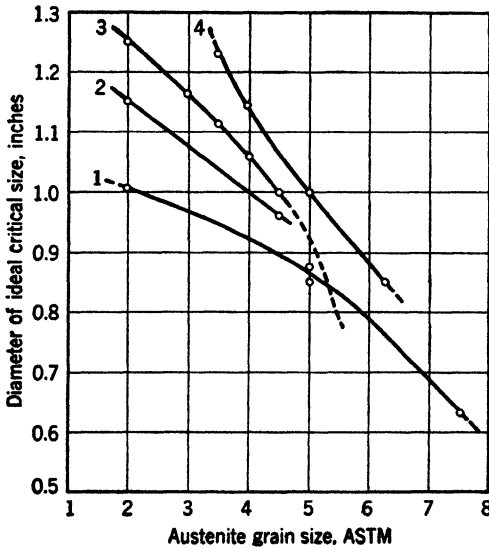


FIG. 211. "Ideal critical size" in four compositions given in Table 28 as changed by austenitic grain size at quenching temperature.

The differences in depth hardening of water-quenched 1-in. rounds are brought out in Fig. 213. Kramer and coworkers<sup>60</sup> point out, however, that Al-killed and Si-killed steels differ in the relations between grain size and hardenability.

**Sluggishness of Coarse-Grained Austenite.** In terms of time required to transform all austenite to pearlite at 1250°, on cooling, when a 1.17 C, 0.30% Mn steel is quenched into a fused bath at that temperature, Davenport and Bain<sup>25</sup> find that all the austenite is

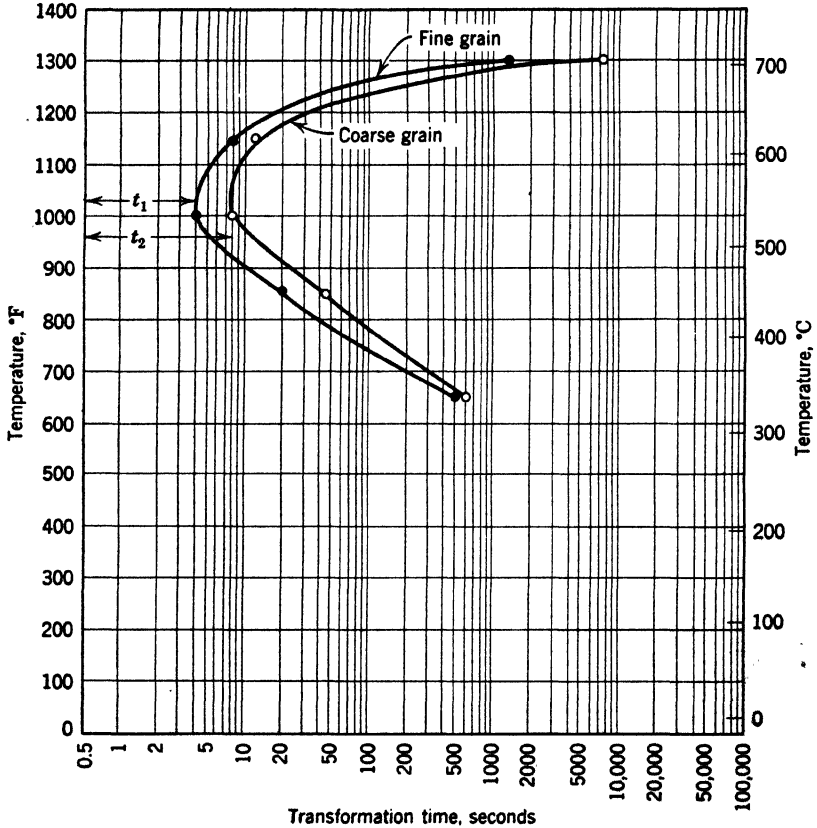
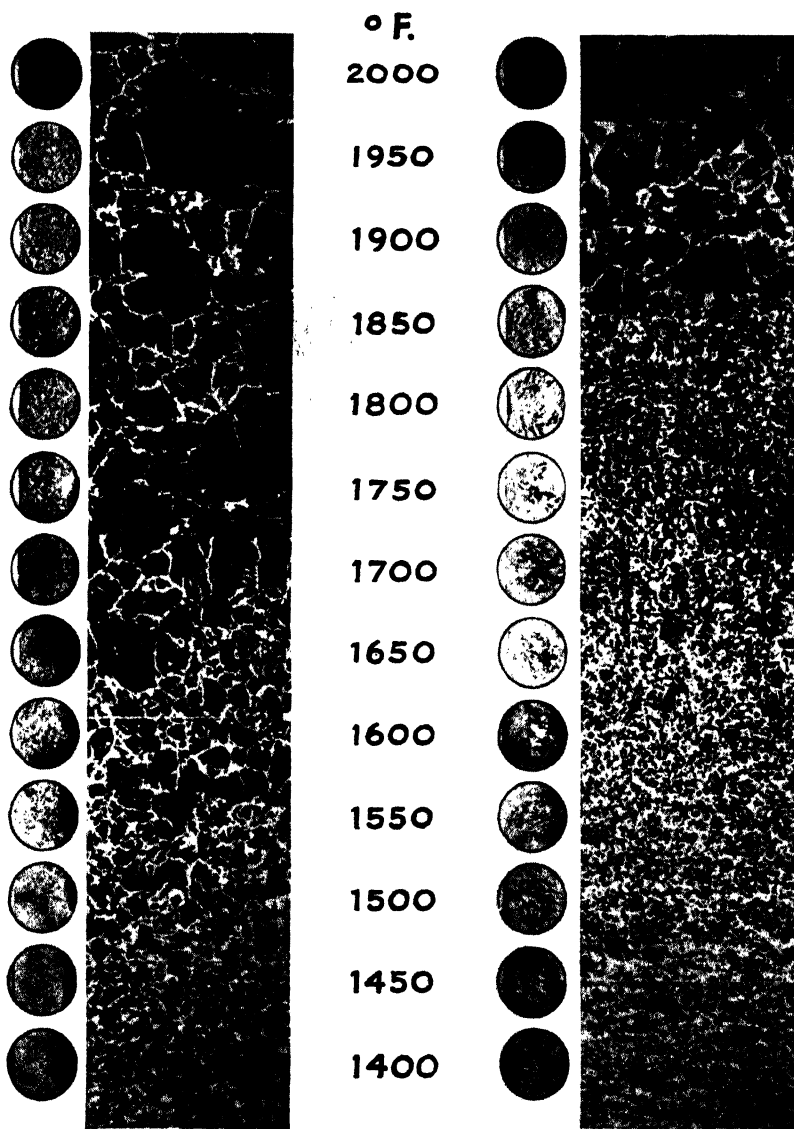


FIG. 212. Influence of austenite grain size on the interval required for completion of isothermal transformation in a 0.9% C steel.

gone in 6 sec when the steel was initially heated to 1550° and had a grain size of 8, that is, was very fine. Coarsening it by heating to 1750°, to a grain size varying from 2 to 6, caused the austenite to take 30 sec to transform completely, that is, 5 times as long.

Thus we see that coarse-grained austenite is sluggish, and much easier to cool down past  $A_r'$ . This makes the coarse-grained steels deeper hardening than the fine. Figure 215 shows Rockwell hardness across cylinders of 1040 steels of the same composition (outside of



(a) Coarse.

0.40 C, 0.71 Mn, 0.18 Si,  
0.018 P, 0.029% S.

(b) Fine.

0.41 C, 0.76 Mn, 0.21 Si,  
0.021 P, 0.035% S.

Fig. 213. Structures and corresponding fractures of coarse- and fine-grained steels, normalized or water-quenched from same range of temperatures. The coarse-grained steel coarsens over the range 1550–1700°; the fine-grained one abruptly at 1900°. The quenched fractures are shown on a larger scale in Fig. 214, the steels are compared as to depth-hardening in Fig. 215, and as to properties of normalized steels in Fig. 216. (Schane)



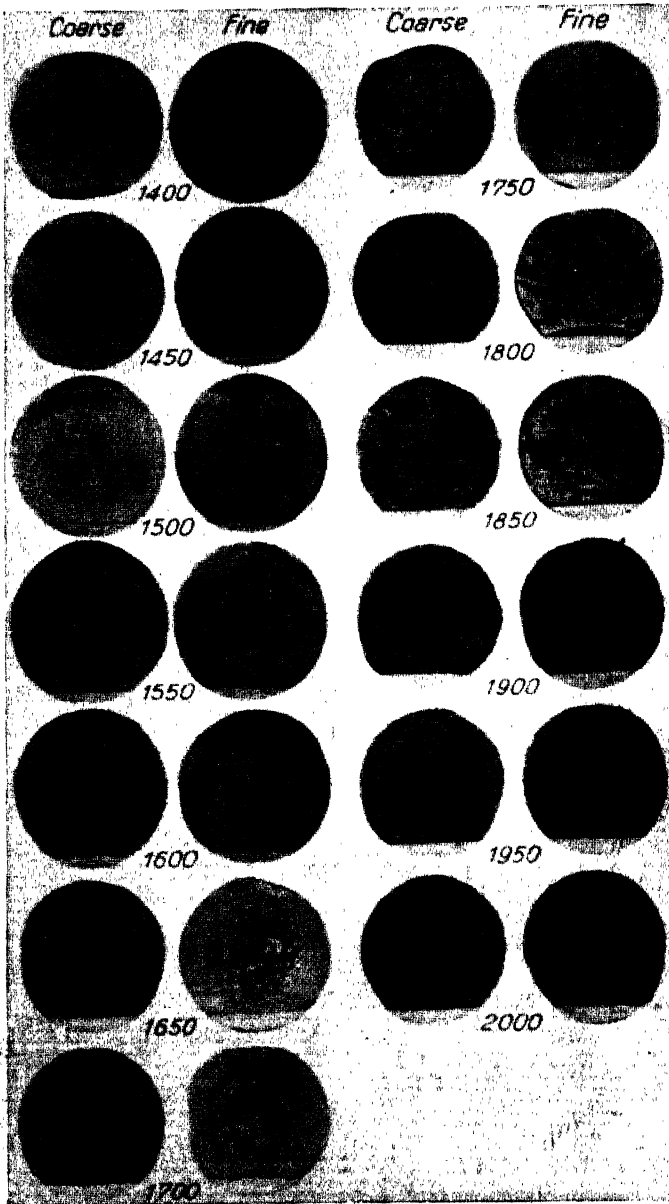


FIG. 214. A larger view of the quenched fractures shown in Fig. 208. (Schane)

their tiny content of Al), but varying in grain size, quenched as shown. The same thing is shown by the depth of the hardened zones on the fractures in Fig. 213.

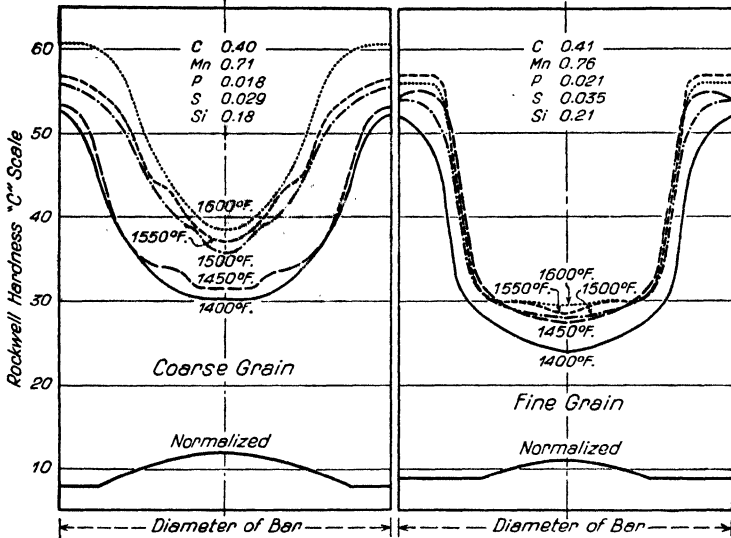


FIG. 215. Hardenability curves on 1-in. rounds of the coarse and fine steels whose structures are shown in Fig. 213, and whose fractures are shown in Fig. 214. (Schane)

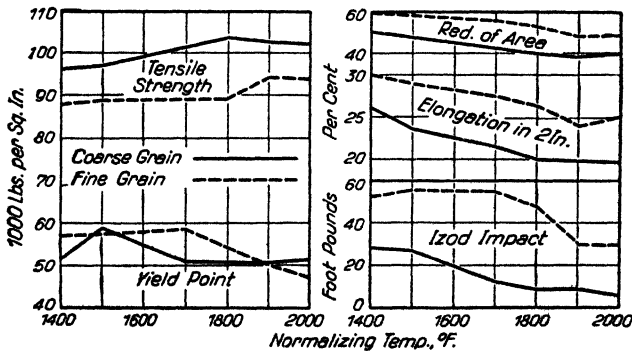


FIG. 216. Mechanical properties of coarse and fine steels, normalized from different temperatures. The compositions are given and structures shown in Fig. 213. (Schane)

**Effects on Quenching Temperature.** Grain size is thus an index of hardenability, and many of the general statements as to critical cooling rates in relation to C content, etc., have to be modified. They

are strictly true only when steels are compared that are of equal grain size at the temperature chosen from which to cool. Thus the effect of the maximum temperature of heating before quenching on

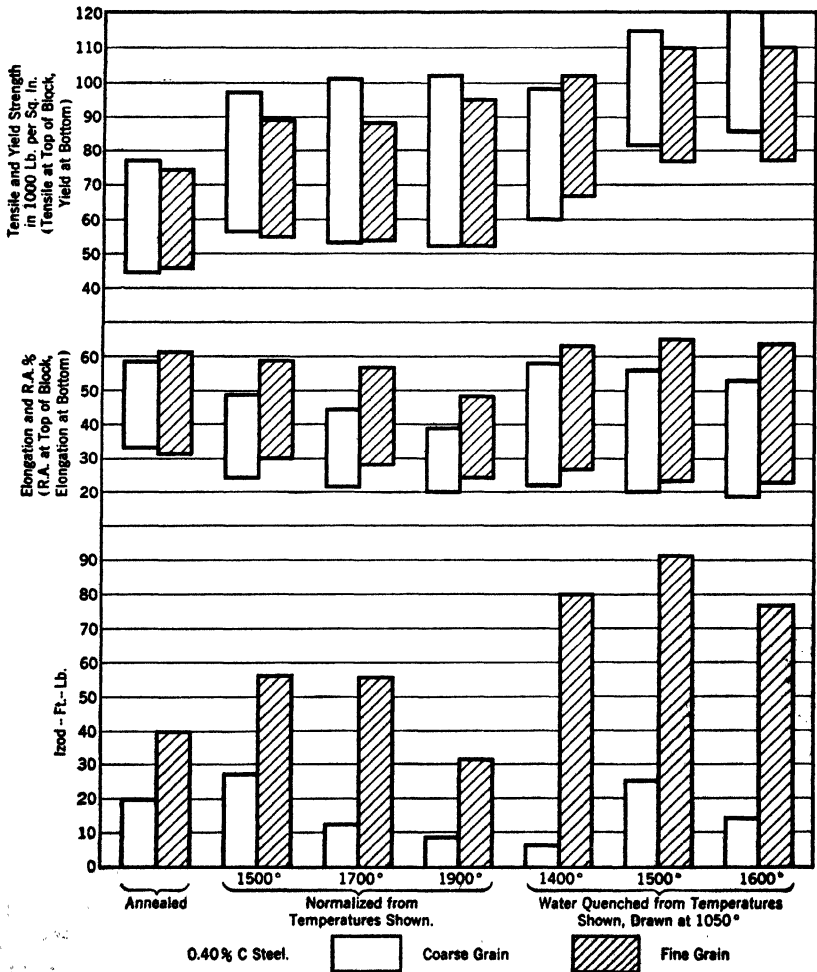


FIG. 217. Effect of grain size on mechanical properties. (After Schane)

the hardenability can be related to the grain size produced by that heating. The repeated warning not to heat far above the end of  $A_{c3}$  for quenching, a sound one for steels that coarsen rapidly just above that temperature, is not necessary and often not advisable for steels that remain fine grained at much higher temperatures. Thus the

fine-grained steels have a much wider safe-hardening range; their heat treatment is less fussy and more fool-proof.

**Effects on Normalizing.** An obvious application for the use of a grain-refining addition, like Al or V, is in steels to be normalized whose critical cooling rate is such that they tend to air-harden. Cases in point are Mn steels with C and Mn both on the high side.

Cast steel that is normalized instead of annealed may show higher strength and lower ductility than is desired. For example, steel of 0.30–0.35 C, 1.50–1.75% Mn, normalized, may show 110,000 tensile with but 15% elong. and 20 ft-lb impact. Suitable addition of V, Al, or Ti ~~so~~ as to make the steel fine grained will drop the tensile to 95,000, raise the elongation to more than 25%, and approximately double the impact. The grain-coarsening effect of Mn and the tendency toward hardening at a low critical cooling rate characteristic of coarse-grained steel are corrected by refining the grain.

An analogous situation is in wrought Mn-bearing steels to be welded. Cooling from the welding heat, in thin sections, tends to produce some air hardening in and adjacent to the weld, with consequent brittleness. Several of the low-alloy high-yield-strength steels (see Vol. III) utilize 1 to 1.40% Mn as chief alloying element, and, in order to secure the desired strength, the C may go up to 0.25–0.30%. Such steels are not considered suitable for welding unless they can be heat-treated after welding. However, the limiting C and Mn contents for safe welding without subsequent heat treatment are still not too clearly defined (see Vol. II). The variations in reported behavior may readily be due to unrecognized differences in grain-coarsening propensities. It is fairly certain that Mn–V steels of this class can be safely welded at a C + Mn content that would be unsafe without the addition of 0.05–0.10% V. Other grain-refining additions would probably have an analogous effect.

**Effects of Grain Size on Mechanical Properties.** Fine-grained hypoeutectoid steels, in cooling through the critical at rates that produce pearlite, do not coalesce the ferrite into a connected network surrounding the pearlite grains so readily as do the coarse-grained steels. A thick connected ferrite network leads to a low impact resistance, its absence to a high one. This, as well as the finer grain in itself, makes the fine-grained normalized steels show high impact resistance, a fact long appreciated in respect to normalized V steels, whose outstanding characteristic is their fine grain.

Data from Schanc,<sup>6</sup> assembled in Figs. 216–217, show that, in whatever condition (annealed, normalized, or quenched and tempered), SAE 1040 is superior in ductility and ~~vastly~~ superior in impact when it is fine grained.

Rolf<sup>26</sup> also gives data for 0.40% C steels, of coarse and of fine grain, water-quenched from 1525° and tempered at 800° to 1200° by 100° intervals. The tensile and yield strengths of the fine-grained steel were 5 to 10% less than for the coarse-grained, but the ductility and toughness were increased. At the lowest draw temperature the elongation and reduction of area were increased 20 to 25%, and the impact resistance was tripled. Similar but less marked superiority in ductility and impact resistance was found at the other tempering temperatures.

Scott<sup>27</sup> finds, for very hard steels, spring temper and harder, that a similar effect holds, as is shown in Fig. 218.

It is universally conceded that, for service at ordinary temperatures, a more desirable combination of strength and toughness is had in fine-grained than in coarse-grained steels. This is most clearly shown by the impact test, and, since the serviceability of the fine-grained steels is universally recognized from actual performance, increased understanding of the effects of grain size has brought an increased use of the impact test.

Bain<sup>28</sup> shows that, whether quenched and tempered or austempered, a coarse structure leads to poor impact (unnotched bars) as indicated in Fig. 219. Since the hardness of all specimens was the same, the effects shown are due to grain size and structure.

The effect on notched-bar impact is also shown in Fig. 220, from the work of McGuire.<sup>29</sup> Here a low-C (0.17 C, 0.44 Mn, 0.21% Si) steel was water-quenched and slightly tempered, in the upper curve, after austenitizing 1 hr at 1600°. The steel had a grain size of 2 after this treatment. To examine what would happen to the core of a carburized piece of this steel, it was put through a "pseudocarburing" treatment, that is, held for 8 hr at 1700° with no carburizer present, and then water-quenched and slightly tempered (lower curve). Under this treatment it grew still coarser grains and, when quenched, transformed largely to lower bainite instead of to pearlite and ferrite.

**Grain Size Perhaps not the Primary Cause.** In spite of all the correlation between grain size and notched-bar behavior, there is reason to believe that it is not grain size itself that governs, but rather a condition, probably incipient precipitation of some sub-

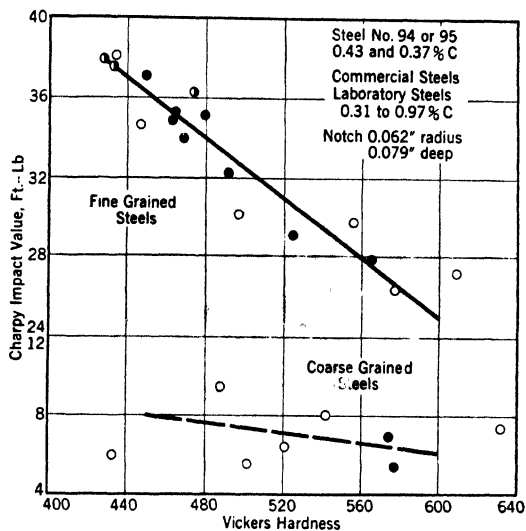


FIG. 218. Impact values of C steels quenched from 1525°, tempered at 300-570°, to hardness shown. (Scott<sup>27</sup>)

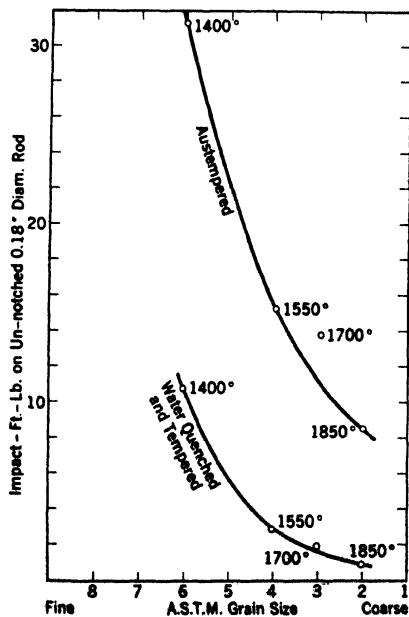


FIG. 219. Grain size versus impact (unnotched). The same steel, containing 0.85 C, 0.42 Mn, 0.18 Si, 0.022 S and 0.008% P, was used in all cases, but was given varying austenitic grain sizes by heating to the different temperatures shown. All specimens were cooled to 1400° before quenching. One series was quenched in molten lead and not drawn (that is, austempered). The other series was water-quenched and then drawn. All specimens had the same hardness, 50 Rockwell C. (Bain<sup>28</sup>)

microscopic constituent, or the opposite, that is, solution of some such constituent, concomitant with grain growth. That is, grain growth may just be a symptom. It is believed possible to have grain growth and retain full toughness, though, in steels as usually made, this very rarely happens. At any rate, before one can justly compare the

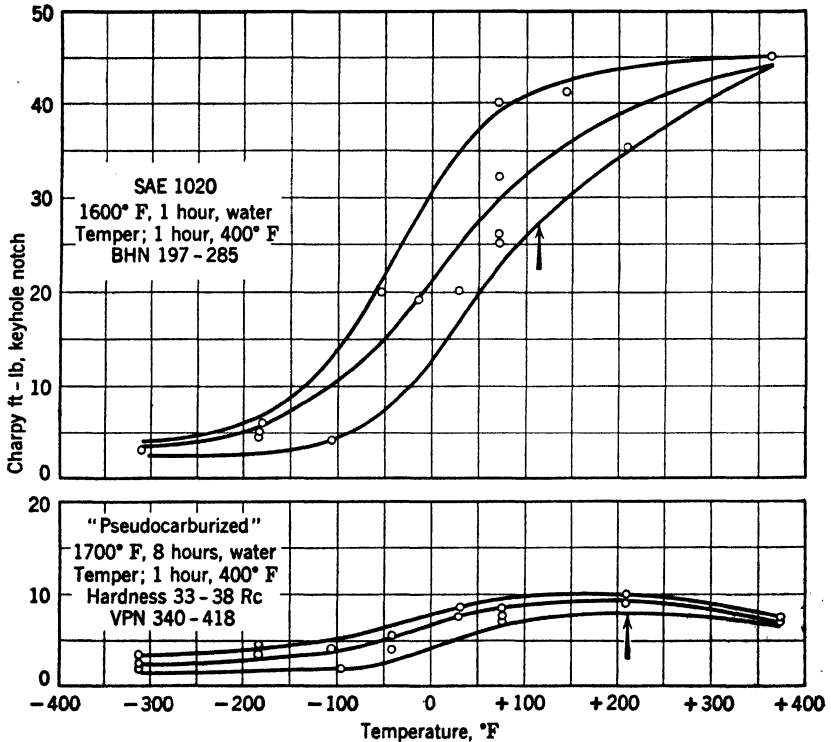


FIG. 220. Upper plot, notched-impact behavior of a coarse-grained 1020; grain size 2. Lower plot, for same steel after long holding at high austenitizing temperature by which still larger grains grow.

effects of two alloying elements or two percentages of the same alloying element in steel, one should first make sure that the steels compared are comparable in respect to grain size and grain-coarsening propensities.

Practically every statement about the heat treatment of and the properties of heat-treated C and alloy steels needs to be modified by a statement of the grain size and the coarsening properties of the steel, or at least to be read while carefully holding in mind the modifying effect of variations in austenitic grain size.

**Grain-Coarsening Propensities versus Creep.** Notwithstanding the close relation between grain size and impact resistance, steels of the same analysis and of the same grain size do not necessarily have the same impact resistance. Data reported by the Joint Committee<sup>80</sup> indicate that at a given grain size in two steels of similar composition, the one initially fine grained and requiring a higher coarsening temperature for grain growth may have a much higher impact resistance than the one with the lower coarsening temperature—as shown in Table 29.

TABLE 29

Steel No.	C	Composition, %			S	Al added, %	Normal-ized from	Grain Size	Coarsening Temp.	Izod Impact	Norm. from	Grain size	Izod
		Mn	Si	P									
54	0.49	0.76	0.20	0.017	0.028	0.025	1650°	2 to 4	1650°	4			
61	0.50	0.75	0.22	0.020	0.026	0.025	1650°	2 to 4	1650°	4			
52	0.46	0.54	0.23	0.016	0.030	0.105	1800°	2 to 4	1625-1775°	18	1550°	6 to 8	27
56	0.51	0.75	0.21	0.018	0.033	0.10	1850°	2 to 4	1700-1800°	16	1550°	6 to 8	23

Steel No.	Norm. from	Tensile	Yield	Elong., %	R.A., %	Grain Size	Creep—Rate of Deformation, % per Hr at End of 500 Hr (unless Noted) at 850° at 20,000 Psi
54	1650°	100,000	49,000	22.5	43.5	2-4	0.00023
61	1650°	107,500	53,500	20.5	40.5	2-4	0.00016
52	1800°	92,000	51,000	25.5	47	2-4	0.00048
52	1550°	91,500	56,000	27.5	51	6-8	0.00910 at 140 hr
56	1850°	106,000	57,500	21.5	42	2-4	0.00019
56	1550°	102,000	59,500	25	49	6-8	0.00415 at 140 hr

As the table shows, a difference of 2 or 3 to 1 in creep rate occurred in steels of the same grain size and similar composition, whereas the difference between the rates of fine- and coarse-grained classes was 20 to 1.

The steels 54 and 61, coarsened to a grain size of 2 to 4 from 1650°, showed only 4 Izod against about 25 Izod for fine-grained steels, but 52 and 56, coarsened to a grain size of 2 to 4 at 1800-1850°, gave around 17 Izod. That is, in the same coarse grain size, the steels of low coarsening temperature lost 80 to 85% of their resistance to impact whereas those that completed their coarsening at higher temperature lost only 30 to 35%.



This is of particular interest because steels with sufficiently coarse grain to show the better resistance to creep at high temperature that accompanies coarse grain size are prone to have very low impact at ordinary temperatures, and a means by which both creep resistance and impact resistance may be maintained at desirable levels would be of great value.

The creep resistance in a pair of lower-C steels, in fine-grained condition, was decreased tenfold by doubling the Al addition, even though the coarsening temperature was not reached and both were normalized from the same temperature to the same grain size.

TABLE 30

	% C	Mn	Si	S	P	Al Added, %	Coarsening Temp.	Actual Normalizing Temp.	Grain Size	Creep—Rate of Deformation at 850° and 15,000 Psi % per Hr
2809	0.29	0.68	0.32	0.045	0.011	0.025	1825°	1650°	8	0.00059 at 500 hr
2810	0.28	0.62	0.31	0.041	0.011	0.05	1875°	1650°	8	0.00580 at 190 hr

	Tensile Properties—Grain Size 8 Norm. from 1650°				Charpy
	Tensile	Yield	Elong., %	R.A., %	
2809	77,000	50,000	32	61	34
2810	76,500	52,500	33	63	37

The steel in which the actual normalizing temperature more closely approached the coarsening temperature gave the better creep result.

**Grain Size an Important Variable.** A duplex structure might have the poor creep properties of its fine grain and the poor impact properties of its coarse grain. It would appear that creep properties in uniformly fine-grained steels may be better, the nearer the fine-grained ones are brought to their coarsening temperature without coarsening (though the creep properties resulting from still higher temperature treatment may be better); and impact properties in uniformly coarse-grained steels may be better, the less the coarsened ones are taken above their temperature for uniform coarsening.

Coarsening propensities, that is, the temperatures of the beginning and end of coarsening and the range of temperature over which duplexing occurs, need consideration as well as mere grain size.

The differences in behavior of coarse- and fine-grained steels at high and low temperatures indicate that they may perhaps be more justly ascribed to "inherent" propensities of the steels, which are *also* indicated by grain size and grain-coarsening temperature, rather than being directly due to the grain size itself. The grain size may be a concomitant rather than a cause.

**Wire and Forgings.** Other features in which coarse- and fine-grained steels differ in behavior are constantly being noted. Boynton<sup>31</sup> prefers coarse-grained wire for cold drawing to high tensile strength, but fine-grained for patenting. Vignos<sup>32</sup> finds that coarse-grained steels make better finished forgings, and Ellis,<sup>33</sup> in a study of forgeability, concludes that, the larger the grain size, the more readily steels are deformed in forging.

**Grain Size in Deep-Drawing Steel.** The importance of grain size in deep-drawing steels is universally recognized. A clear description of the requirements is given by Winlock and Leiter<sup>34</sup> who point out that too coarse a grain size produces rough "orange-peel" surfaces and too fine a one predisposes to stretcher straining. The best balance between the two evils is, they state, found at ASTM grain size 7.

**Grain Size versus Machinability.** Low resistance to impact seems to be connected with easy machinability in *rough* machining. Hence a coarsening of austenite by heating above the coarsening temperature before normalizing is often used to produce a more readily machinable steel; after rough machining the steel may be reheated to a lower temperature at which the austenite does not coarsen, to produce a finally tough steel on the next normalizing or on quenching and tempering. A fine-grained steel gives the smoothest surface on *fine* machining, and so, if we are seeking finish instead of ease of machining, the fine-grained steels are superior.

**Adjustment to Secure Advantages.** By choosing an initially fine-grained steel whose coarsening temperature is not too high, one may make it either fine or coarse grained at will by adjusting the normalizing temperature. Each type has its advantages and disadvantages, well summed up by Wilson<sup>35</sup> as follows: "In general, the coarse grained steels are recommended for good machinability, intense hardening, easy coarsening of structure and high temperature service above 950°. The fine-grained steels are recommended for economical case hardening, good toughness, superficial hardening of the direct hardening steels, small warpage in hardening, best mechanical properties of untreated steel and high temperature service below 950°F."

It is not correct to set 950° as the dividing line in high-temperature service, as this dividing line undoubtedly varies with the composition of the steel, as well as its grain size. In plain C steels the dividing line is probably quite a bit lower, but the general principle of existence of a dividing line of temperature above which coarse grain has superior load-carrying ability seems well established.

**Distortion and Cracking.** Further drawbacks to the coarse-grained steels, and added reasons for not heating to the coarsening temperature before quenching, are the greater distortion on quenching and the greater propensity toward quenching cracks.

With these differences in behavior in heat treating and in properties after heat treating it is obvious that the heat treater can attain his ends not only by selection of a steel of suitable chemical composition but also, and almost equally, by selection of one with the desired grain-size characteristics.

**Juggling the Grain Size.** By choosing a steel whose austenitic coarsening temperature is midway in the feasible range of temperatures for industrial use, he can make it coarse grained or fine grained at will, by heating to a temperature above or below the coarsening temperature. The possibility is also opened up of using controlled-grain-size C steels instead of alloy steels, or of attaining a desired result with a simple alloy steel of controlled grain rather than with a complex or more highly alloyed one.

With the production of steels of controlled-grain-growth tendencies and with the understanding of the phenomena of grain growth, these variations can be utilized with certainty of results. This is a great advance over the situation we were in when the grain size of a heat of steel was an accident and when the heat treater could not be sure that a steel of the same chemical composition as one he had previously used would respond similarly to treatment.

**Speculations as to Mechanism of Grain Growth.** With grain size such an important symptom of differences in behavior, the question arises as to just what is going on at these temperatures in the austenitic range. All that can be seen is grain growth, but submicroscopic changes at the austenite grain boundaries doubtless precede it.

The evidence in the preceding chapter on overheating in forging leads to the impression that some embrittling material accumulates at the grain boundaries, even though the effect is evidenced only by the appearance of the fracture and the tensile data, and not so readily by notched-bar impact.

However, it is generally supposed that grain-boundary material has already been separated at lower temperatures and forms a mechanical barrier to grain growth across these boundaries until it has been dissolved at higher temperatures.

If two sorts of grain-boundary material were postulated, one going into solution at one elevated temperature range, the other being precipitated at a still higher range, all the experimental observations might be reconciled. Since whatever happens is on an invisible scale, recourse must be had to indirect evidence, especially to what the steel maker does in order to control grain-coarsening propensities.

**Methods for Control of Grain Size.** The steel maker<sup>36</sup> is much interested in the methods of control of grain size and the mechanism of the inhibition of growth. Epstein, Nead, and Washburn<sup>37</sup> have discussed the former, and the latter has been very widely discussed, without final agreement being reached on just what the mechanism is. McQuaid<sup>38</sup> lays emphasis on the Al present as metal in Al-treated steels rather than on the aluminum oxide, supposed to be present in submicroscopic form. Brophy and Parker<sup>39</sup> think that both aluminum oxide and excess of metallic aluminum affect the behavior of steel in heat treatment. Some open-hearth superintendents also believe that both are required in producing steels of controlled grain-coarsening characteristics. Foulkes<sup>40</sup> expresses this point of view. Derge, Kommel, and Mehl<sup>41</sup> studied high-purity Fe containing Al, Si, Al<sub>2</sub>O<sub>3</sub>, and SiO<sub>2</sub>, and concluded that as high as 0.20% Al or Si in solid solution did not inhibit grain growth, whereas Al<sub>2</sub>O<sub>3</sub> or SiO<sub>2</sub> did inhibit it. They state: "These results are all in agreement with the hypothesis that fine-grained austenite is the result of inhibition of grain growth by properly dispersed insoluble particles." Most authorities agree with Derge, Kommel, and Mehl in attributing the inhibition of grain growth primarily to mechanical action. In the case of V it is assumed that vanadium carbide is very difficultly soluble and that, until the austenite has dissolved it and attained homogeneity in C content, grain growth does not occur. Solution and homogenization must first be accomplished. If they have been accomplished, for which a temperature well above those ordinarily used for hardening or for normalizing is required, grain growth is rapid. It takes intentional overheating to coarsen such a steel. In the case of alumina or titania and zirconia (and their nitrides) it is seldom realized that these non-metallics have appreciable solubil-

ity, though their solubility may turn out to have an important bearing on the problem.<sup>32</sup>

In fact, evidence is accumulating that  $\text{Al}_2\text{O}_3$  itself is too insoluble to account for grain control and that even coarse-grained steels usually contain as much  $\text{Al}_2\text{O}_3$  as fine-grained ones, for the O in the steel runs about the same and will be combined as  $\text{Al}_2\text{O}_3$  if sufficient Al is added. Nor does the excess metallic Al appear to play a direct part. However, AlN seems to have the proper solubility-temperature relations.

**Probable Mechanism of Inhibiting Action.** The action is considered as a mechanical inhibitor to growth which may appear and disappear according to its solubility relations. The known effect of thoria in W lends color to this view.

Schane<sup>42</sup> points out that the inhibiting substances have some degree of solubility in austenite and the solubility and rate of diffusion vary with temperature. At lower temperatures the substances diffuse to or are rejected to the grain boundaries and prevent growth. At some higher temperatures the boundary film is dissolved or partly dissolved and ruptured so that it no longer prevents grain growth. Dorn and Harder<sup>22</sup> have expanded the idea of solubility of inhibiting materials.

If any of these various kinds of submicroscopic nonmetallic particles are present in the austenite grain boundaries they would act as mechanical barriers to the growing together of adjacent austenite grains, much as an unfluxed oxide film will interfere with welding. At some sufficiently elevated temperature the adjacent grains should be active enough to overstep this barrier, engulf the inhibiting particles, and proceed with grain growth. Actually, at some very elevated temperature this sudden growth (Fig. 209) does occur in the fine-grained steels, in contrast with the more gradual growth of the grains in a steel free from Al, V, or other inhibiting elements.

**Sulphur as a Possible Factor.** Another factor that may enter, as suggested by Sims,<sup>47</sup> is the S content. There is strong evidence that, in cast steels, the type of inclusion formed (Fig. 42) depends on the relative amounts of Al and S, and that some of the Al is combined in the form of an aluminum sulphide. Hence with a given Al addition the Al should divide itself among aluminum oxide, aluminum sulphide, aluminum nitride, and uncombined metallic Al, according to the amounts of O, S, and N present, and the temperature. The aluminum sulphide itself may be one of the nonmetallics that obstruct grain growth, and it would be expected to have a higher solid solu-

bility than  $\text{Al}_2\text{O}_3$ . Titanium and zirconium also form sulphides as well as oxides, carbides, and nitrides. It is quite common to add both Al and Ti for grain-growth control. Halley<sup>49</sup> compares Al, Ti, and Zr. He finds Al most effective at a content of about 0.03% acid-soluble, that is, metallic, Al; Ti very possibly to act through the medium of its carbide; and Zr to be the least effective of the three.

Zirconium changes the distribution of the sulphides much as aluminum does, but titanium does not produce, or at least is not so effective in producing, the distributed type, having a greater tendency to put them into the film type. Hence, TiN is thought to be too difficultly soluble to be an effective grain-growth restrainer by itself. It may be that the purpose of the Al addition is both for control of sulphide distribution and to provide AlN. By combining with part of the O, S, and N, Ti relieves the Al from doing all the work.

In spite of rather good evidence<sup>48</sup> that AlN may be one of the active grain-control agents, since the effective nonmetallic compounds are submicroscopic, it is practically impossible to prove directly just what compounds are present, to say nothing of their exact quantity. For the present, at least, the results can only be described in terms of the amount of addition made to steel at some particular stage of deoxidization that corresponds to a more or less definite content of reactive O, S, N, etc., instead of in quantitative terms of percentage of a certain growth-obstructing constituent.

It might be pointed out that the submicroscopic alumina inclusions postulated to cause or to accompany the agents that cause resistance to coarsening are not necessarily accompanied by the clumps of alumina often found in poorly made Al-treated steels (Fig. 47). The fine-grained steels are not necessarily any "dirtier" than the coarse-grained ones.

**Availability of Grain-Controlled Steels.** The lion's share of all fully killed steels designed for heat treatment is now produced under grain-size control. Even warehouse lots of high-grade steels are usually pedigreed and their grain-coarsening tendencies are on record. Experienced purchasers pay as much attention to the grain-coarsening behavior of a heat as to its chemical composition. The heat treater need no longer contend with unexpected and erratic grain-growth behavior.

**Precipitation Effects.** Grain control by Al, Ti, Zr, V, or combinations thereof, is often combined with hardenability control by a trace of B. Complex addition agents containing several such elements

allow production of steels resistant to coarsening of austenite, with the toughness characteristic of fine-grained steels, yet with the depth-hardening behavior of a coarse-grained steel. When the B addition is made for hardenability, relatively high Mn content, or a low alloy content, as in the N.E. steels, is used, since B has much more effect on an alloy steel, that would show moderate depth hardening even without the B, than on a plain C steel.

These addition elements, used in very small amounts, say 0.01–0.05% (and much less in the case of B) obviously operate through the compounds they form, presumably with C, O, N, and S. (What B is combined with is still in doubt.) The submicroscopic compounds may obviously be dissolved away from the austenite grain boundaries in the course of heating the austenite preparatory for the cooling stage of heat treatment and precipitated during cooling. Whether they are precipitated in the grain boundaries or peppered through the grains will profoundly affect the properties of the steel.

Grain size, grain growth, and their concomitant effects thus appear to be closely allied to the general pattern of precipitation-hardening phenomena. The invisible structure produced by these extremely minute particles, present in very minute amounts, exerts almost as potent an effect on properties as does the visible structure.

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## CHAPTER 16

### FULL ANNEALING

The structures resulting from the transformation of austenite during cooling vary with the temperature at which the transformation starts and the time of sojourn at that temperature. There is a structure characteristic of each temperature level (Chapter 3).

On cooling, hypereutectoid steels first reject cementite, between  $A_{cm}$  and  $A_{r1}$ , and then pearlite; hypoeutectoid steels first reject ferrite, between  $A_{r3}$  and  $A_{r1}$ . If cooling during this rejection is slow, cementite has time to grow into a massive network around the austenite grains or into spines penetrating within the grains of hypereutectoid steels, ferrite has time to collect into large grains, surrounding the pearlite grains. With more rapid cooling the divorce of ferrite is less pronounced; the structure is finer. Slow cooling through the critical range, that is, full annealing, is resorted to in order to produce minimum strength, with near-maximum toughness and ability to stand later cold deformation. It is necessary to exceed  $A_{c3}$  to transform the ferrite wholly to austenite. In steels without any addition for grain-size control, grain coarsening starts at a temperature not far above  $A_{c3}$ , and for such steels an annealing temperature as little above  $A_{c3}$  (Fig. 144) as is feasible with the equipment at hand is always specified when the optimum combination of strength and toughness is desired in wrought steels, since toughness is sacrificed when the austenite has been allowed to coarsen. With steels resistant to grain coarsening, it is only necessary to hold the temperature below that at which coarsening would occur, and, in order to homogenize the austenite in as short a time as possible, annealing at a relatively high temperature is preferable.

### ANNEALING STEEL CASTINGS

Notwithstanding what has been said in the foregoing in relation to holding the temperature not far above  $A_{c3}$  in easily coarsenable wrought steels, a much higher temperature and longer time at tem-

perature are required when annealing steel castings. The original dendritic structure of the ingot has been mechanically broken up in rolling or forging so that in the heating of wrought steel the ferrite patches and the pearlite grains are intermingled and generally made much finer. The austenite grains with 0.83% C formed from the pearlite patches at  $Ac_1$  do not have large ferrite boundaries, and, when  $Ac_3$  is exceeded and the surrounding ferrite has also become austenite but practically C-free, the high-C centers and low-C borders of the austenite grains can equalize their C content fairly readily by diffusion. Moreover, the nonmetallic materials rejected to the grain boundaries of the original ingot have also been distributed throughout the metal by the working.

In cast steels of the ordinary compositions, which have not had the benefit of working, the original coarse-pearlite grains are surrounded by coarse-ferrite envelopes, formed in cooling down after freezing, and the nonmetallic boundary material has not been distributed. To absorb the coarse ferrite, break through the nonmetallic films, and bring about a homogeneous C content in the austenite is a task requiring much higher temperatures and longer times.

Hence, whereas a wrought 0.30% C steel would preferably be heated only around 1550° for annealing, a 0.30% C steel casting will require around 1650°, and the usual prescription is that it be kept at temperature an hour for every inch of diameter or thickness in the largest section before the slow cooling is started.

Normalizing steel castings has uniformly superseded annealing, and quenching, followed by tempering, is rapidly superseding normalizing.

Full annealing involves heating above the critical and slow cooling through the critical range. A time factor is involved. Time and the necessary slow-cooling rate are usually provided by cooling in the furnace, or by burying the piece in lime, ashes, or some other good heat insulator. The softening sought in annealing hypoeutectoid steels is partly brought about by allowing ferrite to separate itself from pearlite instead of being interspersed with pearlite patches. Softening is also favored by getting the pearlite itself to come out in thick layers of ferrite and cementite instead of thin ones. Further softening is accomplished by allowing the cementite layers to break up and contract into balls rather than plates, which process is known as spheroidizing. In hypoeutectoid steels, spheroidization is accomplished *below*  $Ar_3$ ; hence it is essentially a tempering operation and is discussed under that head (Chapter 20) though partial spheroidization can also be accomplished by holding within the  $Ac_3$ - $Ac_1$  range.

When the initial structure is very coarse, as may result from over-heating in forging (Chapter 14), especially in large sections, a single anneal may not refine the grain sufficiently. Treatment in two or more stages may have to be resorted to. The first anneal results in some refinement, distributing pearlite patches in the ferrite matrix more uniformly, so that homogenization is facilitated on the second reheating into the austenite range.

## S CURVES AND ANNEALING

The traditional full anneal calls for slow cooling in the furnace. To retard and control the cooling some heat can be supplied to the

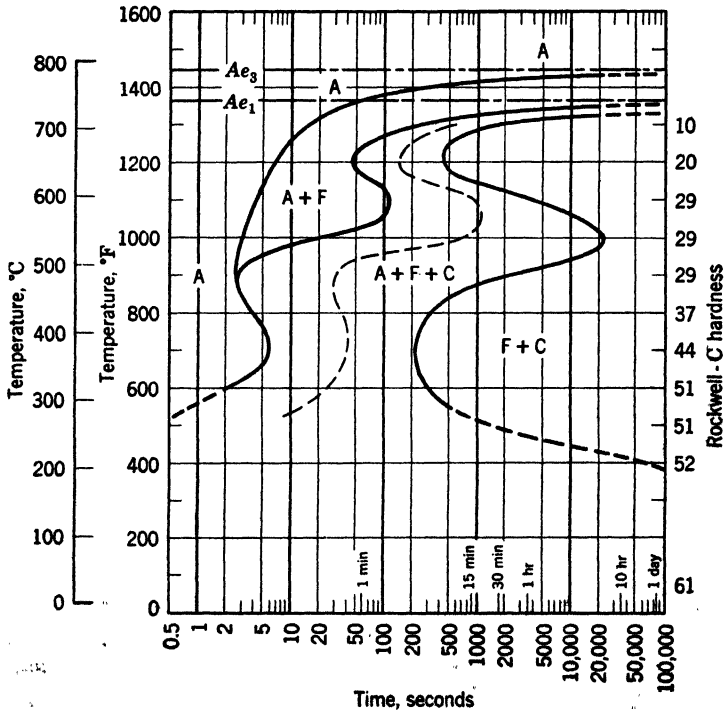


FIG. 221. SAE 4140 containing 0.37 C, 0.77 Mn, 0.98 Cr, 0.21% Mo, austenitized at 1550°F; grain size: 7-8. (U. S. Steel Corp.'s *Atlas of Isothermal Transformation Diagrams*)

furnace. When the stored heat in the piece will suffice, it can be cooled in lime, but then there is no control. Using the annealing furnace for cooling ties up a high-temperature furnace. Obviously,

transferring the piece to a second furnace operating at lower temperature may often be economical. Then we want to know at what temperature to run the second furnace. After the austenite has all transformed, nothing is gained by continuing the slow cooling.

Some insight as to the requisite temperatures and times can be gained from the S curves, by noting the ending time and the hardness produced by holding at different temperature levels. Thus Fig. 221 for 4140 shows that *under isothermal conditions* a 3-hr sojourn at 1300° is required to allow the transformation to be completed, and this softens to 10 RC.

Only some 10 min at 1200° is required for completion, but, due to the lower temperature, the hardness is raised to 20 RC. Dropping to 1100° prolongs the time to about 1 hr, and raises the hardness to 29 RC. If 20 RC is acceptable, "quenching" to and holding at 1200° would hasten the operation. Running a second furnace at 1200° and transferring the piece to that furnace would then seem better than running it at 1100°.

For sections thin enough to cool rapidly to 1200°, the scheme works as the isothermal S curve indicates. For heavy sections a complication comes into evidence, as shown by the work of Liedholm and Coons, already mentioned (Chapter 12), in that the separation of ferrite in cooling from 1425 to 1200° enriches the untransformed austenite in C, and, when that higher-C austenite transforms, the transformation products are harder. When the austenite is enriched by 0.15% C, what was 4140 now becomes, locally, 4155, and 4155 isothermally transformed at 1300° has 22 RC; at 1200°, 31 RC. The short-cut method hence may turn out less attractive in practice than it is in theory.

## CHAPTER 17

### NORMALIZING

**Normalizing.** In the discussion of annealing it has been pointed out that, in order to secure ready absorption of excess ferrite and equalization of the C in austenite, it is necessary or advisable with steel castings, overheated forgings, and very large forgings, to heat to much higher temperatures than in regular annealing, even though in so doing the austenite grain size may become very large. To avoid the coarse structure that would result from the slow cooling of regular annealing, more rapid cooling, in the open air, is employed. Similarly, in C tool steels and in case-carburized steels a preliminary heating to a high temperature, above  $A_{cm}$ , may be advisable to break up the continuity of the cementite network surrounding the pearlite, and air cooling is used since slow cooling would re-establish the network.

The same procedure of air cooling may be used even on steels that would readily homogenize at ordinary annealing temperatures, when a finer structure is desired than that resulting from slower cooling. When the more rapid cooling is to be used, so that a fine structure will result anyway, it is not so important to avoid the grain growth of austenite that comes from higher heating, but it is desirable to attain homogenization of the austenite. So homogenization is secured by heating to a higher temperature for a shorter time.

The rate of cooling of austenite that has been given a good degree of homogenization profoundly affects the structure. Confining ourselves in this chapter to such rates of cooling that austenite all decomposes to ferrite and cementite while passing through  $A_{r1}$  and none is retained to change into martensite (that is, there is no quench hardening), we find that changes in the cooling rate will alter the way the ferrite and cementite are deposited as they form from the austenite. Thus different structures and different properties may be obtained.

**Cooling Rates.** Cooling rates are generally vaguely described as "fast" or "slow," or by mentioning the way they are produced, for

example, by furnace cooling; cooling in ashes, in still air, in an air blast; quenching in oil, in water, or in a high-pressure water spray. These phrases have to be further clarified by stating the mass of the steel being cooled. A very fine wire will lose heat faster in ashes than the center of a huge forging would under the most drastic water quenching. The really important matter is the *rate* at which the steel cools through the critical range, and this can be expressed in degrees per minute. Unfortunately, the measurement of cooling rates is not easily carried out; so the qualitative phraseology ordinarily has to be employed. We have some idea of the speed of travel when we use the words "at a walk," "double-quick time," automobile driving speed, aircraft cruising speed, speed of a bullet, although each of these varies, and we know the true speed only when it is definitely expressed, as in miles per hour.

A few determinations of cooling rates are on record. Howe, Foley, and Winlock<sup>1</sup> found that very slow furnace cooling is at the rate of about 2° per minute, and, in air cooling of small pieces in an air blast, the cooling rate is around 120° per minute. To harden a 0.2% C steel to the maximum of which it is capable requires that it cool faster than about 2400° per second, just as it passes through the critical.

**Mechanism of Separation.** In a hypoeutectoid steel, as austenite cools below  $A_{r_3}$  it starts to deposit its excess ferrite, and this proceeds until all excess ferrite has been rejected around patches of austenite that have finally become enriched in C to the eutectoid concentration. If the cooling from  $A_{r_3}$  to  $A_{r_1}$  is slow, the ferrite has time to grow into a connected or coalesced network with islands of 0.83% C austenite inside. If the cooling is fast, the patches do not so readily grow together and remain peppered more uniformly throughout the mass. At  $A_{r_1}$  the C-bearing austenite has to break up into ferrite and cementite.

If the cooling at  $A_{r_1}$  and just below it is extremely slow, as in a spheroidizing anneal, the cementite tends to contract, by something analogous to surface tension, into balls within a matrix of connected ferrite (Fig. 37). In any but this excessively slow cooling the austenite splits up into alternate layers of ferrite and cementite, the lamellar structure of Figs. 4 and 35. But the thickness of each layer is less, the more rapid the cooling. Hence in furnace cooling we get coarsely lamellar pearlite, easily resolved into the layers at rather low magnifications. With more rapid cooling the plates become thinner and thinner, more of them packed into the same space, but the



plates are still quite flat. Patches of pearlite have the eutectoid ratio, about 0.83% C, on very slow cooling; but rapid cooling gives pearlite of other ratios. It is possible to have pearlite (in hypoeutectoid steels) with as low as, say, 0.60% C. (In such cases the estimate of the C from the amount of pearlite shown on a micrograph is not accurate.) Still more rapid cooling gives not only still thinner plates, but warped ones, so that their cross section in a micrograph looks curly. Still faster cooling brings the warpage and the thinness of the plates to such a degree that, instead of layers, the microscope shows a confused, emulsified mass. This emulsified stage was called, in the older nomenclature, a "sorbitic structure." A still further stage, in which dark-etching clumps appear, has been called "primary troostite," or "troostite of quenching," since it appears when quench-hardening rates of cooling are nearly approached (Fig. 27). This structure is now classed as bainite.

**The Term "Sorbite" Reserved for Quenched and Tempered Structures.** Modern microscopic study has shown these latter stages to be nothing more or less than the ferrite and cementite of pearlite, in non-lamellar form. Vilella, Guellich, and Bain<sup>2</sup> suggest that the terms sorbite and troostite should not be used for the structures resulting directly from decomposition of austenite, but should be reserved for the decomposition structures of martensite on tempering.

Under this nomenclature, after austenite in cooling becomes of eutectoid or near-eutectoid composition, it decomposes to pearlite or bainite (unless it is retained to go on down to martensite, or even retained as austenite itself, under the influence of the rapid cooling rates of quenching).

The stages then range, with increased cooling rates, from coarse pearlite (course in respect to thickness of lamellae), through fine pearlite, on to curly pearlite, and finally to bainite.

When the cooling is fast enough to produce the curly or the fully emulsified pearlite, it is also fast enough to avoid coalescence of ferrite into a network, and the ferrite is jumbled up with the pearlite.

**Properties of Pearlite.** The continued subdivision makes increasingly for strength, and, although ductility may be a trifle lower than with coarse pearlite and coalesced ferrite, it is still high.

The strength varies with the width of the ferrite and cementite plates in the pearlite, as Belaiew<sup>3</sup> and Greene<sup>4</sup> have shown. If they are about 0.4 micron \* wide, in a eutectoid steel, it shows 100,000

\* A micron is  $\frac{1}{1000}$  of a millimeter, that is, 0.000039 in.

tensile. If they are 0.25 micron wide, the strength rises to 150,000. As the width of the plates decreases, the warping of the plates into the curly structure goes on, finally reaching the bainite stage with a tensile of some 225,000.

Hence the fine pearlite which is obtained by air cooling of small specimens of C steel, or whatever approach to it we can get in larger specimens, is valuable as a structural material. The heat treatment to produce it is short; hence normalizing, that is, heating above  $A_{c_3}$  and air cooling, is much employed. It is used as a prelude to full annealing or to quenching, in order to put the steel into such state that the ferrite does not have to look very far away for the C that is coming into it from the pearlite on a reheating to form homogeneous austenite preparatory to the final cooling, be that cooling slow or rapid. It is used also by itself, without any further re-creation of austenite, for hypoeutectoid steels. For hypereutectoid steels it is generally used only as a prelude.

Normalizing is applied to the thinnest and the thickest pieces of hypoeutectoid steel we have to heat-treat.

**Normalizing of Sheet Steel.** Very low-C unalloyed steel is used for tin plate, automobile body sheet, and the like. Such steel will run around 0.10% C, or below, too low for effective hardening by ordinary quenching means. Such steel must be very ductile, since ductility is required in forming seams on tin cans or in stamping an automobile top or fender. Strength is usually of little relative importance, but the forming properties are all-important.

In modern practice the thin sheets are hot rolled to perhaps  $\frac{1}{16}$  in. singly, or in continuous mills as strip, then either pack-rolled hot, or cold-rolled continuously to final gage. Even the pack-rolled material has some rolling strains and needs to be annealed, whereas the cold-rolled sheet must be softened in some way.

The maximum of ductility and formability is obtained by a full anneal; but, since grain growth and resultant roughness of the formed surface ("orange peel") come from a full anneal, one is restricted to annealing below the critical, or to normalizing. Severely cold-rolled strip, because of its strained and deformed structure, softens on annealing at 1200°, whereas pack-rolled material requires higher temperatures, sometimes up to 1400°. In either case in the regular annealing cycle, stacks of sheets or coils of strip, often in lots of 10 to 50 tons, are covered with a box or bell into which a neutral or nonoxidizing gas atmosphere is led (as is more fully discussed in Vol. II) and slowly heated to the annealing temperature by electric resistors, by

radiant tubes, or by direct firing outside the box. Very slow heating is required to avoid overheating of the edges because of the poor thermal conductivity of the piles or coils, and, for the same reason, they cool very slowly. The heating and soaking cycle is usually some 20 to 40 hr, and it takes another 24 hr or more to cool to a low enough temperature, about 280°, so that the charge will not oxidize when the box is opened. Keelan<sup>5</sup> has discussed the schedules for strip annealing in detail.

In contrast to this, single sheets or continuous strip, can be raised to a normalizing temperature of 1750° in 1½ min, held 1 or 2 min, cooled to 800°, reheated to 1150° for 1 or 2 min, and then cooled, all in a space of less than 15 min. Or, a cycle can be used of 1½ min, heating to 1750°, holding 1 min, cooling in 1½ min to 1020°, retarding the cooling so that it takes 2 min to drop to 840°, and then cooling to 280° in 3 min.<sup>6</sup>

The reheating to 1150° or the retarded cooling from 1020 to 840° in this case is probably not employed for quite the same reason that steels of higher-C content are reheated or drawn after normalizing but more probably in order to control precipitation-hardening effects (see Chapter 10). Ferrite has a very small solubility for cementite, so that at temperatures above the solubility line *OP* of Fig. 135, some of the cementite that came out at the grain boundaries on the first cooling can be redissolved on reheating above this line but below *A<sub>r1</sub>* and by controlled cooling can be precipitated within the grains instead. Other elements, such as O<sub>2</sub> and N<sub>2</sub>, present in very small amounts in these unkilld rimming steels, have rather similar solubility curves in ferrite to that shown in Fig. 135, and the treatment will probably redissolve them all.

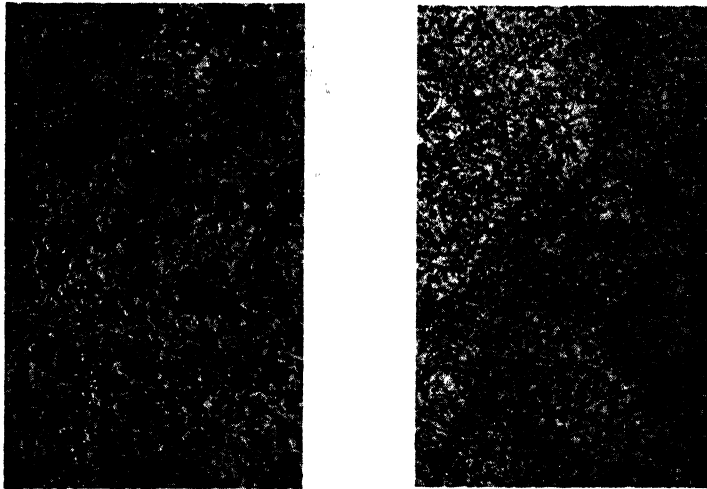
Such normalizing gives formability almost equivalent to that of box-annealed materials, with fine grain and hence good surface on the stampings. The same steel, normalized and box-annealed, is shown in Fig. 17.

**Normalizing of Steel Castings.** Although full annealing, that is, slow cooling from a high temperature, has been the usual treatment for C-steel castings, normalizing is now more in vogue. The air cooling of fairly heavy castings, particularly when a whole carful is pulled out of the oven in which they were heated, is rather slow, but yet enough more rapid than a furnace cooling to give different properties.

Sims<sup>7</sup> cites an example of two structures obtained from the same heat of cast medium Mn-V steel, shown in Fig. 222. Note the uni-

formity in tensile properties as contrasted with the improvement in impact with the finer-grained steel.

The importance of sufficient heating prior to normalizing to homogenize the steel is brought out by Fig. 223, also from Sims. The raw structure in a cast 1-in. cube was somewhat refined after 1 hr at 1650°, but it took 6 hr at 1650° to obliterate the raw structure in the



(a)		(b)	
			× 100. (Sims 7)
Tensile	98,000	101,500	
Yield	63,000	66,000	
Elong. %	27	27.5	
R.A. %	55.5	57.5	
Isod, ft-lb	18.5	48	

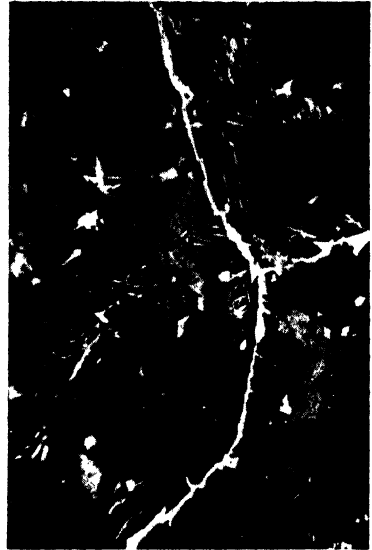
FIG. 222. Structure versus impact resistance of two specimens from the same heat of Mn-V steel. Both specimens were double-normalized, but the second normalizing on (a) was from an excessively high temperature.

center of a cast 6-in. cube. The grains in the latter are still coarse, and a second normalizing would be required to produce a structure like that of the 1-in. cube.

**Normalizing of Forgings.** In heavy forgings, such as locomotive driving axles and the like, one would like to develop the maximum combination of properties, but quenching to secure it can hardly be used on C steels, for the objects are so massive that C steel will not harden throughout. Moreover, unless the piece is symmetrical, it is likely to crack on quenching. In forgings for big guns alloy steels are employed that will harden much more deeply on quenching, so that the guns may be quenched and tempered. When the cost of the steel and of the treatment is such a factor as to eliminate this, and C



(a) Center of 1-in. cube, as cast.  
× 100.



(b) Center of 6-in. cube, as cast.  
× 25.



(c) Same as (a) after 1650°, 1 hr, air.  
× 100.



(d) Same as (b) after 1650°, 1 hr, air.  
× 100.

FIG. 223. Normalizing of light and heavy sections of the same heat of cast steel.  
(Sims<sup>7</sup>)

or moderately alloyed steels that would not harden throughout are used, these huge forgings are either full annealed as discussed in Chapter 16 or they may merely be normalized.

Fry<sup>8</sup> has discussed locomotive forgings of 0.34% C steel in a manner that clarifies the question of structure and properties versus rate of cooling in normalizing. Forgings 12 in. in diameter by 3 ft long were left solid, or bored out as was shown in Fig. 18, heated together to 1500°, held there 6 hr, and air-cooled. The more rapid cooling as the section is decreased improves the structure and the ductility.

**Tempering after Normalizing.** The normalizing treatment is usually followed by a tempering, below the critical, to remove cooling stresses.

No apparent structural change occurs on such tempering, but more or less marked improvement in ductility results. Data which show this, kindly supplied by the International Nickel Company, are given in Table 31.

TABLE 31

FORGED STEEL—0.30 C, 1.20 Mn, 1.60% Ni

	Normalized from 1550°	Normalized 1550° and Tempered 1000°
Tensile	94,000	94,000
Yield	73,000	71,000
Elong., %	27	29.5
R.A., %	51.5	58

CAST STEEL—0.29 C, 0.74 Mn, 3.57% Ni

	Annealed	Normalized 1650°	Normalized 1650°, Tempered 1200°	Double Normalized 1650°, Tempered 1200°
Tensile	103,000	111,800	103,000	102,250
Yield	65,750	80,750	71,000	74,500
Elong., %	22.5	12.5	22	24
R.A., %	41.5	15	46	47

CAST STEEL—0.35 C, 0.60 Mn, 3.55% Ni

	Annealed	Normalized 1650°	Normalized 1650°, Tempered 1200°	Double Normalized 1650°, Tempered 1200°
Tensile	104,000	114,700	104,500	105,000
Yield	64,500	72,500	71,500	73,250
Elong., %	22	12	22.5	23.5
R.A., %	42	19	45	50

Although this set of data does not show the changes in impact, they are likely to be even more marked than the changes in static ductility. Cast steels with less of a tendency toward air hardening than the Ni-Mn steels are not so notably improved by tempering after normalizing.

In terms of the S curve, normalizing for production of fine pearlite, with its higher strength than an annealed, or a coarse pearlitic structure, requires that the transformation of austenite be delayed to a temperature well below 1335°. That is, austenite for normalizing needs to be somewhat sluggish, with the pearlite nose of the S curve lowered. The time delay needed to permit evading the nose in quenching to produce martensite is, however, not required for normalizing. Hence, steels to be heat-treated by normalizing do not require alloying for sluggishness to the same degree nor in the same sense as do those for quenching and tempering.

Another difference in the two classes of steels appears in that alloy content has very little influence on the properties of quenched and tempered martensite; the alloy was there primarily to aid in the production of martensite, not to affect its properties. Normalized steels are, however, materially influenced by the amount and kind of alloy contained.

Alloyed ferrite, with strengthening elements in solid solution in it, is appreciably superior to unalloyed ferrite. Hence the properties of normalized steels are influenced not only by the amount and distribution of the carbide, but also by the composition of the ferrite matrix, whereas, in steels quenched to martensite and then tempered, the carbide size and distribution has so much more potent an effect as to swamp out the effects of any differences in the ferrite.

The steels especially designed for normalizing ordinarily contain only a low-alloy content. They form a class of "low-alloy high-yield strength" steels or "mild-alloy" steels which are discussed in Vol. III. It happens that this class shows considerable similarity to the N.E. steels, and, as Stewart and Wiley<sup>9</sup> show, some of the N.E. steels, although primarily designed for quench hardening, are nicely fitted for normalizing in large sections.

Stewart and Wiley<sup>9</sup> show, even more strikingly than does the table, p. 411, the benefits of tempering these normalized N.E. steels. The effect is primarily one of stress relief, not alteration in structure.

**Stress Relief of Normalized Steel.** The elimination of stress is of particular importance in the case of rotors for steam turbines. The

forgings may be as large as 5 ft in diameter by 20 ft long, and, since, for efficiency, the turbine is designed with very small clearances, any appreciable warpage or growth would cause rubbing of the moving parts. The permanence of dimensions must be maintained over the life of the turbine, 20 years or more. The turbines operate at high temperatures, with superheated steam under high pressure.

Stress relief, like other heat-treatment processes, is a matter of both time and temperature, and so the operating conditions offer opportunity for gradual warpage to relieve any stress that is left in the forging. The choice of the steel, and its treatment, must be calculated to minimize the retention of stress.

Mochel<sup>10</sup> points out that the stress introduced in normalizing will depend on the temperature of the *A<sub>r</sub>* range and on the stiffness or plasticity of the steel at that temperature. Comparing a 0.42% C steel having its transformation on cooling at about 1250°, with a 0.25 C, 2.70 Ni, 0.30 Mo, 0.10% V steel in which the transformation is not completed until the temperature has fallen to about 815°, and recalling (see Vol. III) that Mo increases creep resistance at elevated temperatures, that is, the steel is stiffer, he concludes that the stresses in normalized forgings of the latter are much the greater and demand very long stress-relieving treatments. Even the C steel, with its greater plasticity, requires long treatment as is evidenced by the following program for treatment of 0.42% C rotor forgings of 40-in. diameter and larger.

The forging is not allowed to cool, but immediately after forging is transferred to a hot furnace and held at 1600° for 30 hr, and then air-cooled to black. It is then reheated to 1200° and held there for another 30 hr, followed by furnace cooling.

Rough machining is then done, and the forging is placed in a cold furnace, slowly heated to 1480°, held 27 hr, again air-cooled to black, reheated to 1200°, held 30 hr, and furnace-cooled. This reheating is done with the forging suspended with its axis vertical in a tall furnace.

Finish machining is then done, and a final stress-relieving draw is given, again with the forging vertical, at 1050°, holding 48 hr at temperature and furnace-cooling. That is, the rotors are double-normalized and given three stress-relieving draws.

**Normalizing versus Annealing.** Although the ductility shown in a tensile test may be greater with full annealing than with normalizing, the higher yield and tensile strengths accompanying the more rapidly cooled structure give a greater area under the stress-strain curve of the tensile test, that is, greater ability to absorb work. Coupled



with this is the specific effect of the finer structure, the combination leading to greater toughness in the impact test. Howe and co-workers<sup>1, 11</sup> demonstrated this nicely.

As Table 32 shows, the beneficial effect of the normalizing rate of cooling as compared with annealing is more evident on forging grades than on steels of higher-C content.

TABLE 32

INFLUENCE OF VARIATIONS IN SLOW RATES OF COOLING ON THE MECHANICAL PROPERTIES OF CARBON STEELS

C, %	Cooling Rate, °F per sec	Tensile	Yield	Elong., %	R.A., %	Charpy Impact, Ft-Lb	Brinell Hardness
0.34	0.02	70,000	35,000	28	50	6	130
	2	78,000	45,000	34	60	22	150
0.52	0.02	85,000	42,000	27	38	4	160
	2	95,000	53,000	28	42	8	190
0.75	0.02	90,000	40,000	20	35	1.8	180
	2	120,000	55,000	16	28	2	230

**Stability of Structure.** Normalizing and then tempering below the critical but well above the temperature of service is a favorite treatment for carbon and low-alloy steels for use at high temperatures where great stability of structure is desired. The creep resistance of steel with lamellar pearlite is much above that of the same steel with its cementite spheroidized. Bailey<sup>12</sup> has emphasized this and has shown that spheroidization may occur well below the critical temperature if the steel is held at temperature many thousands of hours. Thus a creep test of the steel for a thousand hours or so, in which the steel still retains its original structure, leads to the expectation of a certain order of creep resistance, but, if on very long heating the structure shifts toward the spheroidized condition, the creep resistance falls. The stability of the original structure is an important thing to know in evaluating steel for high-temperature service. The change should not be unexpected, since, like any other structural change on tempering, spheroidization is a time-temperature function.

In ordinary heat treatment the time that can be given is at best relatively short, so that the obvious variable is temperature. In long-time service the time factor has a chance to show itself.

Although the theoretical explanation is not obvious, it is found that a high-temperature draw below the critical, but well above the temperature of service, tends to stabilize the structure against spheroidization at the lower temperature. A normalize and draw, although producing lower room-temperature strength than a quench and draw, usually gives much better high-temperature stability and continued effective creep resistance.

Tempering after air cooling is, of course, applied also to steels that air-harden, that is, those whose critical cooling rates for formation of martensite are so lowered by alloying that an air cool is actually a quench. Air hardening is not properly called normalizing, because it does not put the steel in the "normal" pearlitic state.

**Effect of Cooling Rate after Welding.** Even though air hardening, in the sense of the production of martensite, may not occur, C steel of more than 0.15% C, and still more notably, low-alloy steels with more than that amount of C, may be hardened, that is, very fine pearlite may be produced, when the zone next to a weld cools from the welding heat. When a welded specimen shows any zone over 200 Brinell, a stress-relieving draw may be used to prevent having a brittle joint. Large welded pipes are stress-relieved by being annealed at around 1200°. Special induction heaters have been designed, hinged to allow their being placed around the completed weld, so that the joint may be annealed in place after welding.

When steel is cut by a torch, the metal adjacent to the part that has been melted out will cool rapidly and become hardened if the C or alloy content of the steel is sufficiently high. Flame-cutting outfits are often supplied with a lower-temperature diffuse flame which follows the cutting flame and makes the metal adjacent to the cut cool more slowly.

**Patenting.** As was mentioned on p. 245, a rate of cooling designed to give pearlitic structure is applied to wire to be further fabricated by cold drawing. The wire, usually of rather high C, from 0.60 to 0.85% C, is heated to a high temperature and a certain degree of grain growth is intentionally produced. Temperatures may be as high as 1850° or even 1925°, but the cooling is at such a rate as will give very fine pearlite. This rate of cooling can be obtained in air on the usual sizes of wire, but a steel with a slightly greater propensity toward hardening, that is, one with Mn raised above the

usual range, is usually chosen when air cooling is to be applied. A quench in hot lead held at around 900 to 1150° provides the proper rate of cooling and a more controllable one than air cooling. The structure obtained is then broken down by the wire drawing, the original grains drawing out into a fibered structure in the finished wire. The properties of lead-patented cold-drawn wire are slightly superior to those of air-patented cold-drawn wire. Either the air- or the lead-cooled wire is better for this particular purpose (of being cold-drawn) than wire of much smaller grain size produced by quenching to martensite and tempering back to sorbite. Lead patenting is a variety of austempering.

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## CHAPTER 18

### AUSTEMPERING, BAINITE, SLACK QUENCHING

In Chapter 12 S curves were discussed with special attention to their limitations in respect to hardenability. A type of heat treatment for which the S curves do have direct applicability (and limitations) may now be discussed. This is the "austempering" process, in which austenite is transformed to other products without intermediate production of martensite. It might be considered as akin to normalizing. In normalizing, a controlled pearlite structure, and in "austempering" (the commercial application of the isothermal holding used in getting S curves), a controlled bainite structure, is sought. The structures, and hence the properties, vary with the temperature at which transformation is forced to take place.

Forcing it to take place at the desired temperature in heavy sections is more easily said than done, as has been discussed previously (p. 36). Hence, in trying to connect up structure and properties with transformation temperatures, tiny specimens have to be used, since from the consideration of cooling rates, it is clear that only small sections can be trusted to have surface and center behavior nearly enough alike so that the structure will be uniform and that the properties represent just one structure.

**Behavior of Nonuniform Structures.** Before considering uniform structure, it is of interest to note some results by Smith, cited by Grossmann<sup>1</sup> on nonuniform structures obtained by isothermal treatment at high temperature, where first ferrite, then ferrite plus pearlite, separate; but for times insufficient for complete transformation of austenite. He used a 3140 steel, of 0.38 C, 0.74 Mn, 0.27 Si, 1.39 Ni, 0.76% Cr, austenitized at 2100° and transferred to a salt bath at 1240°. After different holding times, the partly transformed specimens were water-quenched, and thus martensite was produced from the remaining austenite. The martensite was then slightly tempered (400° for 1 hr). Notches (not standard) were then ground into Charpy impact specimens, and impact tests were made.

An initial all-martensite structure, so tempered, gave 20 to 40 ft-lb. When time was allowed for ferrite to form at 1240° before quench-

ing, considerable ferrite or ferrite plus pearlite had formed, and the specimen was then quenched and tempered, the energy absorbed fell to 5 ft-lb, despite the presence of these supposedly much tougher patches in the tempered-martensite matrix. The ferrite was restrained against plastic flow by the surrounding martensite. When

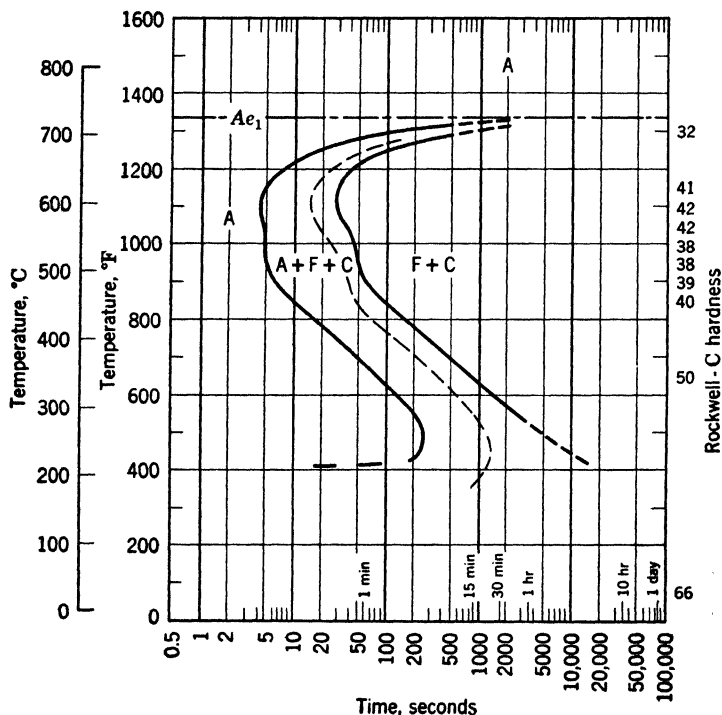


FIG. 224. S curve for a eutectoid steel, containing 0.79 C, 0.76% Mn, austenitized at 1650°F; grain size: 6. (U. S. Steel Corp.'s *Atlas of Isothermal Transformation Diagrams*)

the amount of ferrite plus pearlite increased to 75%, so that the tempered martensite was enveloped, and flow could occur in the ferrite; the energy absorbed rose to 40 ft-lb. The first ferrite to form was of the blocky type shown in Fig. 23.

**Behavior of Uniform Structures.** Gensamer and coworkers<sup>2</sup> have studied nearly eutectoid steels with various Mn contents, using specimens treated in 0.26 in. diameter, and tested as 0.25-in.-diameter 1-in. gage length. The small size was chosen in order to produce uniform structures.

The properties found represent those corresponding to the end of the S curve, that is, after completion of transformation in the pearlite and high-temperature bainite regions. One of the steels is very close

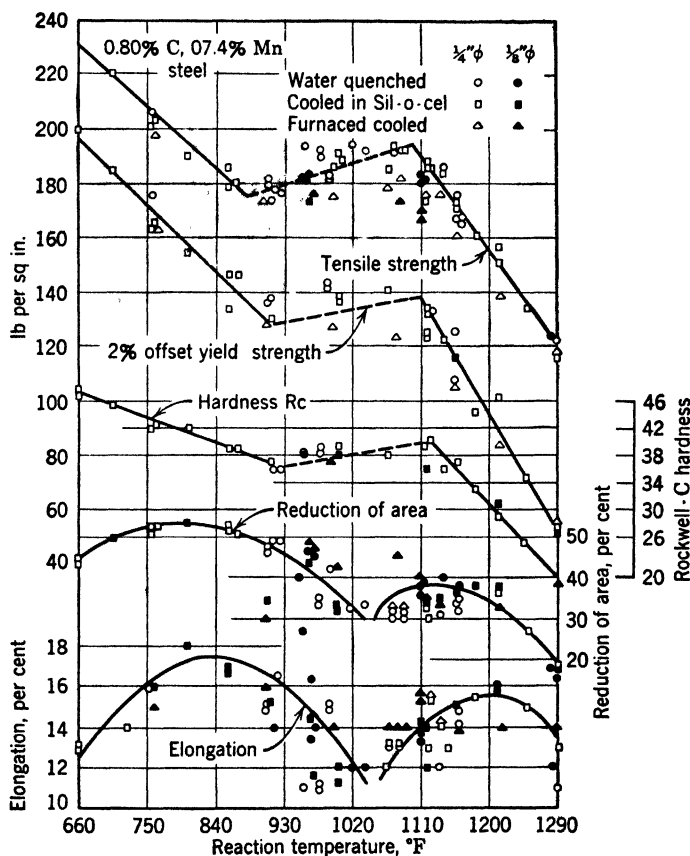


FIG. 225. Mechanical properties versus mean reaction temperature, plain C eutectoid steel. Open points are for  $\frac{3}{4}$ -in. specimens; filled-in points for  $\frac{1}{8}$ -in. specimens; circles are water-quenched after reaction; squares are cooled in Sil-o-cel; and triangles are furnace-cooled.

in composition to one for which an S curve (Fig. 224) is available, the chief differences being that the steel of the S curve was austenitized at  $1650^{\circ}$  and had a grain size of 6, whereas Gensamer's was austenitized at  $1560^{\circ}$ , and its grain size was 5.

These workers found a linear drop in tensile strength from 225,000 at  $375^{\circ}$  to 130,000 at  $1250^{\circ}$ . With the transformation temperature at  $1300^{\circ}$  the strength dropped to 110,000. The ductility showed a

peak around 1065°, which coincides with the nose of the pearlite separation part of the S curve, and with the smallest spacing of ferrite and pearlite lamellae in a regular rather than warped arrangement.

From Fig. 225, for eutectoid steel and from the results for other steels, Gensamer concludes that pearlite and bainite are quite different products, even though there is no break in the S curve to mark a dividing line. In Fig. 225, the ductility curve shows two branches.

The pearlite range covers hardness from 20 to 40 R.C. The properties resulting from different reaction temperatures are shown in Table 33. The optimum combination of strength and toughness for

TABLE 33

Temperature	R.C.	Tensile	Yield	Elong., %	R.A., %
1300°	20	120,000	50,000	12	20
1150	35	170,000	120,000	15	35
1065	38	180,000	130,000	11	30

pearlite comes at the middle of the temperature range. The yield ratio increases greatly as the spacing decreases. Holding at these temperatures might be considered a refined type of normalizing. Holding at lower temperatures is the "austempering" process.

**Spacing in Pearlite.** The spacing of the carbide layers in the pearlite produced at each temperature was determined for the four

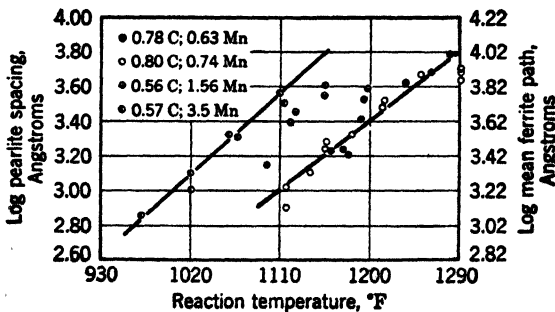


FIG. 226. Relation of spacing in pearlite to reaction temperature.

steels used, and is shown in Fig. 226. Gensamer stresses that the mechanical properties are related to the length of the path that can be traversed in the structure by dodging around the carbide and going through ferrite only, until carbide bars the way. The carbide spacing and, hence, the length of the ferrite path, decrease with the tem-

perature at which the transformation occurs. Increasing C and Mn contents, as Fig. 225 shows, also decreases the spacing. Spacing is controlled by the rate of diffusion of C in austenite.

**Limiting Sizes for Austempering.** The 0.26 in. diameter in which these specimens were transformed in the hot-bath quenching allowed even the center to retain austenite for transition to bainite at the selected temperature. With steels of lower C or Mn contents, in which the pearlite nose is nearer the left edge of the S curve, that is, the pearlite time delay is shorter, the successful formation of bainite, free from pearlite, is only obtained at the center of still smaller pieces.

Legge<sup>3</sup> gives the limiting sizes shown in Table 34 for various compositions. These diameters correspond to the center-cooling rates

TABLE 34

C, %	Mn, %	<i>Approx. Max. Diam.</i>
		<i>in In.</i>
1.00	0.40	0.15
1.00	0.75	0.19
0.85	0.40	0.155
0.85	0.75	0.22
0.65	0.75	0.19
0.65	1.10	0.28
0.65	1.80	0.63

that just miss the pearlite nose of the true S curve on quenching into a fused bath held at about 600°.

The austempering process is not recommended for C or low-alloy steels under 0.55% C; the pearlite nose juts out too far in lower-C steel.

High-alloy steels, of lower C content, in which the pearlite nose is far back, can be used, but in such steels attention has to be paid to the time for *completion* of the bainite transformation in order that the austempering operation shall not take too long to be practical. If one wanted to get 50 R.C., 4140 would take 10 min at 600°; 4340, 1 hr at 525°; and 4360, some 6 hr at 575°, according to the S curves. The maximum hardness at which all bainite can be produced is limited by the temperature of martensite formation.

Daash<sup>4</sup> gives the results of 15 min holding at various hot-bath temperatures on 1045 and 3130, using 0.505 in.-diameter bars. The data are shown in Fig. 227. As would be expected, the size was too



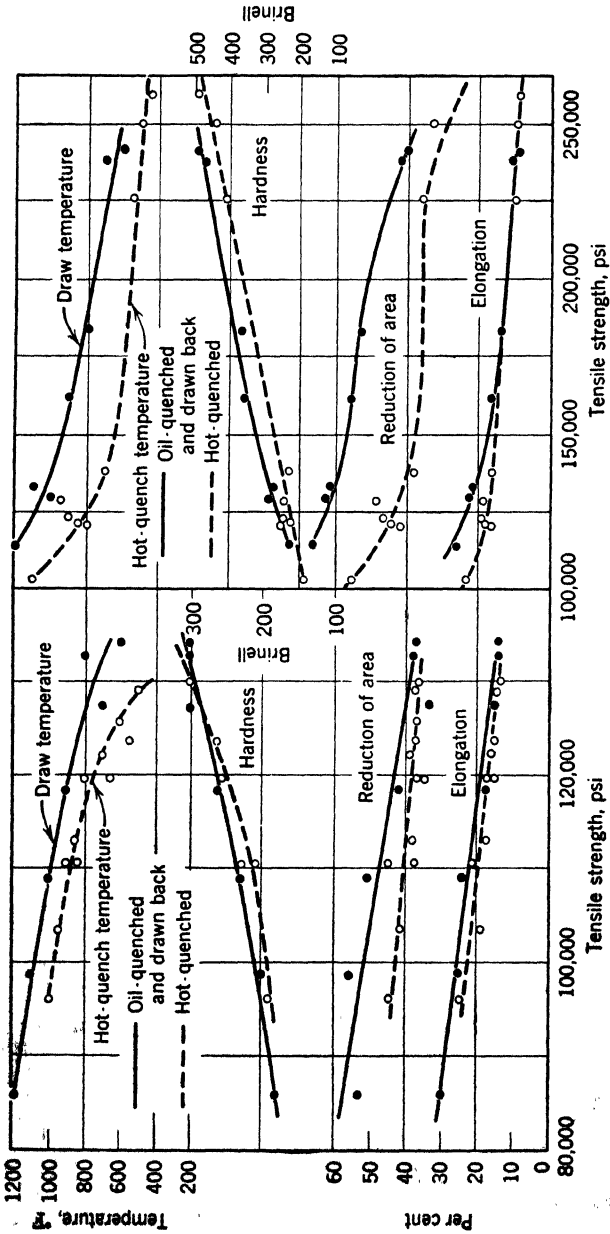


FIG. 227. At left is a comparison (on basis of equal tensile strength) of 0.505 X 2-in. test bars of 1045 oil-quenched and drawn versus quenched in hot bath. At right are similar data for Ni-Cr steel 3130.

great for austempering 1045; only pearlitic structures were obtained, as is shown by the hardness at the lower hot-bath temperatures; 3130, however, produced bainitic hardness. For a given tensile strength, the R.A. was lower for the bainitic structures. Yield strengths were not determined.

**Optimum Hardness Range.** The useful hardness range on austempered steels of this type is around 40 to 58 R.C., and the optimum properties are met, for each particular steel, within a smaller range, of about 4 points R.C. within this range, since in either the harder or softer bainitic structures, the properties are less good than when the steel is quenched and tempered to that hardness. Within that limited hardness range, corresponding to low draw temperatures and low ductility in the case of tempered martensite, the combination of strength and toughness in the austempered product is often considered superior, especially when a battering type of wear is to be encountered, because of greater toughness. For example, a 0.78% C, 0.58% Mn steel, in 0.18-in.-diameter rod austenitized at 1450° gave the results shown in Table 35.

TABLE 35

	<i>Regular Oil Quench, Tem- pered at 650° for ½ Hr</i>	<i>Quenched into Fused Bath at 600°, Held 20 Min</i>
Hardness, R.C.	50	50
Tensile, psi	259,000	259,000
Elong., %	3¾	5
R.A., %	26	24
	Free bend of ¼ in. length broke at 45°	Bent 150° without breaking

At higher draw temperatures for the oil-quenched and higher austempering temperatures, balanced to produce equal hardness, this steel gave the results shown in Table 36. Another steel of 0.85% C,

TABLE 36

Hardness, R.C.	45	45
Tensile, psi	250,000	254,000
Elong., %	6	6
R.A., %	42½	49
Hardness, R.C.	41	41
Tensile, psi	226,000	225,000
Elong., %	9	9
R.A., %	42½	54

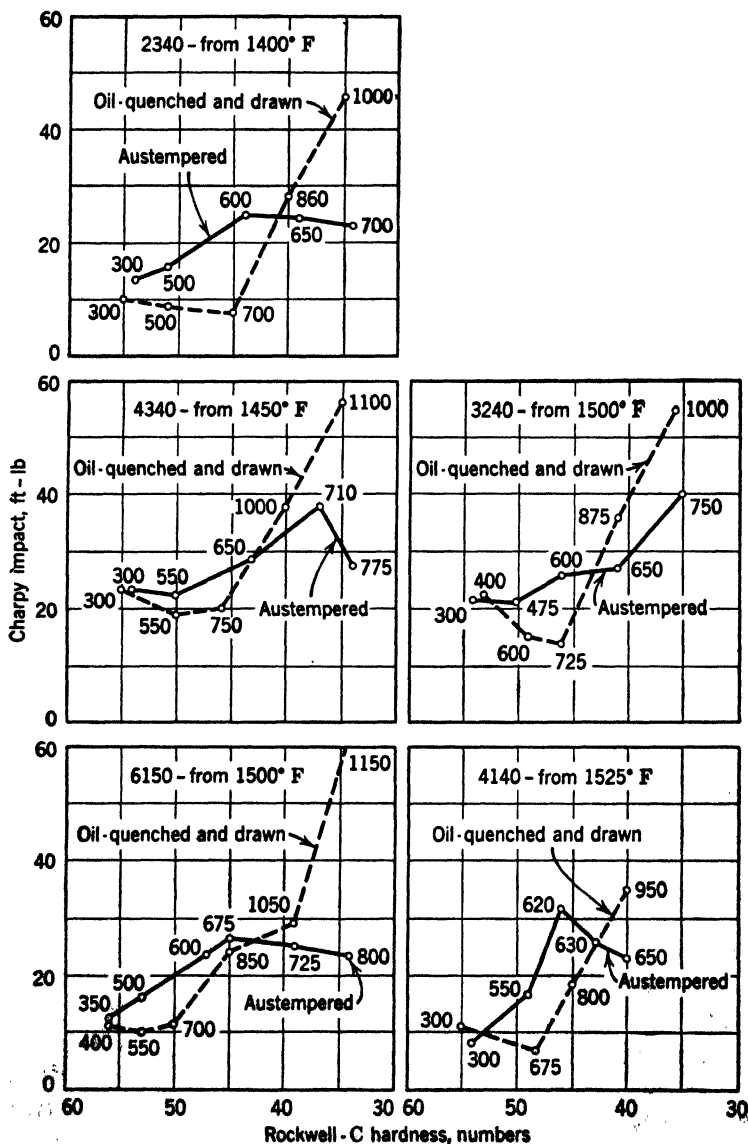


FIG. 228. Relation between Charpy impact and hardness numbers after austempering 1 hr at temperatures noted (followed by air cooling) and after quenching in oil and drawing back 1 hr at temperatures noted (followed by air cooling). All samples were held 45 min at the heating temperature before they were quenched.

0.42% Mn, also tested in 0.15 in. diameter, gave the results shown in Table 37.

TABLE 37

Hardness	50	50
Tensile, psi	261,000	257,000
Yield, psi, 0.1 set	225,000	211,000
Elong., %	4½	6
R.A., %	28	45

This set of figures reveals that the yield ratio of the austempered specimen is lower than that of the quenched and tempered one, which explains why austempering is not applied to springs in which high yield strength and freedom from "set" are sought. Zimmerli<sup>17</sup> comments on the low fatigue strength of austempered springs. Legge remarks that, for equal yield strengths, austempered stock must be a point or two higher in R.C. hardness than the same steel conventionally treated.

The limited hardness range in which austempering has an advantage is shown in Fig. 228 for room-temperature tests on several steels by Payson and Hodapp.<sup>5</sup>

**Impact Behavior.** The greater plasticity at a high hardness level makes the austempered stock stand more of a bend without rupture and may raise the impact resistance on unnotched specimens when austempering has produced the optimum hardness. Legge shows some spectacular impact data for unnotched specimens, whereas with conventionally notched specimens the results are scarcely distinguishable from those on quenched and tempered ones. At 50 R.C., 2340 treated in 0.4 × 0.4 in., austenitized at 1425°, oil-quenched and tempered at 500°, or austempered 4 hr at 500°, compared in conventional key-hole notch Charpy tests<sup>6</sup> as shown in Table 38. However, when

TABLE 38

Temperature of Impact Test	Quenched and Tempered, Ft-Lb	Austempered, Ft-Lb
Room Temperature	14-16	15-20
-40°	10-15	15-20
-120	8-11	10-15
-175	8-10	9-12
-310	5-6	6-7

the treatments were oil quenching and tempering was at 875° or austempering 1 hr at 700° to 35 R.C., the figures were those given in

Table 39. When the hardness was reduced to 22 R.C., the two foot-pound-temperature curves were parallel, but with that for austempering some 10 ft-lb lower than that for quenching and tempering, over the whole temperature range.

TABLE 39

Room Temperature	30-33	30-38
-40°	26-30	22-30
-120	22-27	15-22
-175	19-22	9-14
-310	13-18	5-6

A plausible explanation for the somewhat better toughness of austempered stock, in its optimum high-hardness range, over that of quenched and tempered stock is that martensitic quenching is liable to produce cracks in the martensite needles, as has been shown;<sup>7</sup> and it takes a rather high tempering temperature to heal the cracks. Because there is not such a marked volume change in the formation of bainite, and its formation is more gradual, internal cracks seem to be absent from properly austempered stock, and also in that in which martensite forms gradually, not all in a rush. Nevertheless, the martensite formed extremely rapidly in inductive surface hardening is claimed not to have internal cracks.

**Bainitic Cores.** In the case of the slack-quenched interior of a piece that is martensite on the outside and bainite on the inside, the bainite is not of one hardness level, as it is in small pieces properly austempered, but has a hardness gradient, the softer center not having optimum properties. In the tempering of the piece, both the martensite and the bainite are tempered. The properties of tempered martensite are well known, those of tempered bainite are much less known. It would take proper austempering of small pieces, followed by tempering and determination of properties, to establish the behavior of tempered bainite from an initial bainite structure that was the same throughout the piece. The data of Hollomon and co-workers<sup>15</sup> help to fill this gap.

Liedholm's data<sup>8</sup> (Fig. 176) showed that bainitic structures temper more rapidly than martensitic structures. Engel<sup>9</sup> studied the tempering of martensite, bainite, and fine pearlite, giving the curves of Figs. 229-230, which show the consistently lower hardness level of bainite in the tempered state. Mechanical properties were not determined by Engel. Greaves and Rees<sup>10</sup> oil-quenched a 0.35 C, 1.20

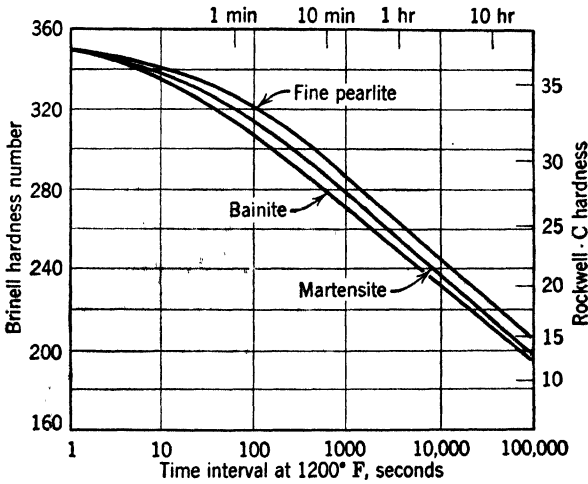


FIG. 229. The softening, with increase of long-time interval, at 1200°F of three characteristic structures all of same initial hardness. (Engel)

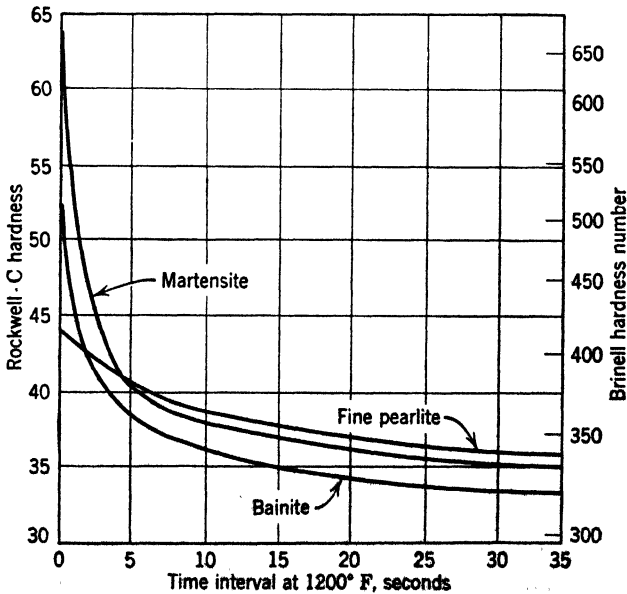


FIG. 230. The softening, with increase of short-time interval, at 1200°F in three characteristic structures of C steel. (Engel)

Cr, 0.57 Mn, 0.26% Mo steel from 1545° in 1 in. and 3 in. diameters. The over-all quenched hardness of the 1-in. section was about 57 R.C., that is, chiefly martensite, that of the center inch of the 3-in. bar, about 43 R.C., that is, bainite. After tempering at 1200°, tensile

TABLE 40

	1-In. Bar	Center of 3-In. Bar
Brinell	289	251
Tensile, psi	134,500	119,500
Yield, psi	114,500	94,000
Elong., %	21	21
R.A., %	64	58

specimens from the two gave the results shown in Table 40. For approximately the same ductility, the strength of the tempered bainite is lower.

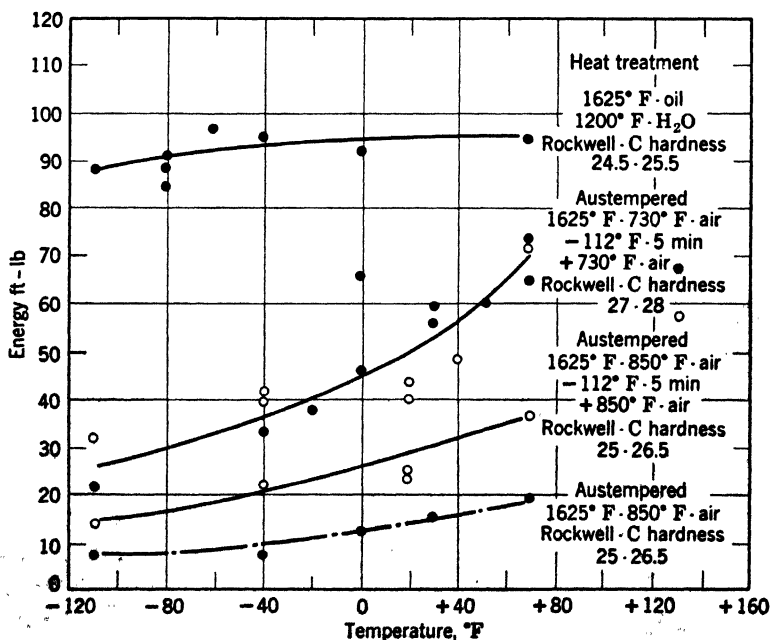


FIG. 231. 4130 steel, Charpy impact, V notch.

A coarse-grained 1045 water- and oil-quenched steel in 0.4 × 0.4 in. and tempered at 900° to around 32 R.C. gave the low-temperature key-hole notch Charpy curves of Fig. 112. The slack-quenched tempered structure is definitely inferior to the fully quenched tempered structure.

Watertown Arsenal impact data, cited by McGuire<sup>6</sup> for 4130 in  $0.4 \times 0.4$ -in. size, quenched and tempered, and austempered at  $730^\circ$  or  $850^\circ$  (in some cases given cold treatment in the hope of transforming any retained austenite, followed by tempering at the austempering temperature), are shown in Fig. 231. At the 25 Rockwell-C hardness level, the austempered specimens were much inferior.

Shepherd<sup>11</sup> says, "The maximum combination of physical properties can be obtained only by the use of structures that are approximately 100% martensite as hardened, that is to say, structures containing no other transformation products. Many published tests showing the relationship between mechanical properties and tempering temperature give a very wrong impression. Tests of this character are only good for the particular percentage of martensite that existed in the tensile test section as quenched."

**Slack Quenching.** The inferiority of a tempered slack-quenched structure compared with a tempered fully-quenched structure is well appreciated by ordnance experts. Maximum yield strength for an acceptable toughness level is demanded in big guns. Projectiles may be designed so that the sharp tip will dig into the armor on an oblique hit, and then shatter or crack off, leaving the base intact to act as a punch, or so that the whole projectile will pass through intact. High-speed photographs of a projectile hitting a plate show that the projectile is plastically deformed before it enters; it punches a bigger hole than its own undeformed diameter, and recovered projectiles show the results of plastic deformation as they twisted their way through the armor. The armor itself is expected to be gouged and dented, and even to be fully penetrated by a projectile with sufficient "oomph," but it is not expected to show back spall. Also, good behavior of guns, projectiles, and armor is desired not only at ordinary but also at subzero temperatures.

For withstanding such conditions of service, the tempered product from the slack-quenched bainitic structure is decidedly inferior to that from a fully martensitic structure. Laboratory data indicating this are given by Rosenthal and Manning.<sup>16</sup>

That does not mean that equal inferiority exists when the service of a heat-treated piece does not involve plastic deformation or static notch toughness. More commonly, practical service depends primarily on static yield strength, on surface hardness, or on fatigue resistance. In fatigue, where the nucleus for failure is the prime criterion, the surface must resist crack formation. The rate of propa-



gation of the crack beneath the surface is of interest, but experience with carburized parts shows that a sharp crack, once started at the surface, propagates so fast through even the toughest core that the properties of the core make little difference. Hence, for most cases of service resistance against fatigue, there is little demand on the core for anything beyond sufficient hardness to support the still harder surface against cracking due to plastic deformation of the core beneath it.

But where surface hardness and strength are not enough (and the piece should have as good yield strength at the center as at the surface) slack quenching fails to give the needed yield strength. All the cases where high center strength is ordinarily demanded are certainly not justified by the engineering requirements of the service, but it is often justifiably demanded. In such cases, alloying to make austenite sluggish enough to transform to martensite without bainite must be resorted to for heat treatment in large sizes.

Despite the good toughness of untempered bainite in the optimum hardness range, the fact that on either side of the range the properties are inferior limits the applicability of austempering to those very thin sections in which the necessary cooling rate can be produced on quenching into a hot bath. Austempering makes a very good shovel blade, but very poor heavy armor.

In slack-quenched structures, the bainite is not all of optimum hardness; ordinarily some is above and some below it. Moreover, such structures, accompanied by exterior martensite, are tempered, since the martensite must be tempered, and so in slack quenching we are dealing with tempered bainite, not bainite in its best condition.

Reid<sup>12</sup> brings this out in comparing aircraft-engine cylinders from two lots of 4140 of almost identical chemical composition: 0.42 C, 0.74 Mn, 0.96 Cr, 0.23% Mo, quenched in still oil and in oil at high pressure (the latter, of course, giving more martensitic, less bainitic structure), then tempered to 296 Brinell (32 R.C.), as shown in Table 41. The still oil gave lower strength than would be expected at this hardness.

TABLE 41

Quench	Still Oil		Oil at High Pressure	
	Steel 1	Steel 2	Steel 1	Steel 2
Tensile, psi	109,600	117,700	151,000	151,000
Yield, psi	101,000	105,000	135,000	136,000
Elong., %	3	5½	10	8½
B.A., %	5½	8	15	15

**Yield Ratio.** Elmendorf<sup>13</sup> produced some martensite in 0.66, 0.78% Mn; 0.80, 0.87% Mn; 0.91 C, 0.46% Mn; and 0.64 C, 0.61 Mn, 0.25% Mo steels by "time quenching," that is, allowing definite percentages of martensite to form from austenite quenched from 1550° by regulating the temperature of the quenching bath (oil at 80 to 320°, salt at 450 and 500°) and the time it operates. After the requisite number of seconds, the piece was transferred to a higher temperature bath (530 or 700°) and held in it for a few seconds to temper the martensite (to darken it for metallographic examination so the per-

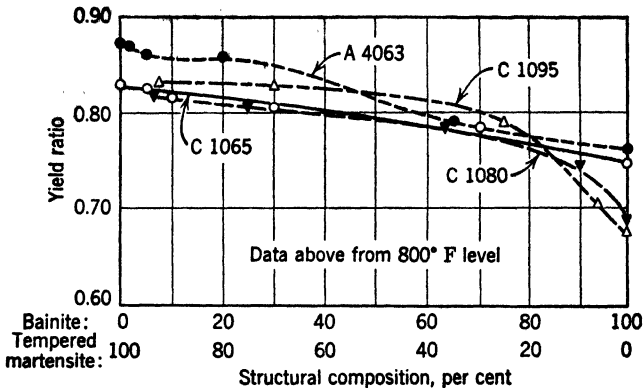


FIG. 232. Effect of structural composition on the yield ratio of four commercial steels.

centage of martensite and remaining austenite could be determined). A similarly quenched specimen was then austempered at 600 to 900° by 100° steps. The previously formed martensite was, of course, simultaneously tempered. The decreased yield ratio as the amount of martensite produced on quenching decreases is shown in Fig. 232.

Elmendorf points out that in these steels bainite under 45 R.C. has unsatisfactory properties. His aim was to produce a mixed structure of tempered martensite and bainite of a high hardness level with toughness increased over that of tempered martensite. This was accomplished for the 0.80% C steel (typical of all four) as Fig. 233 shows. However, it is significant that he used specimens of 0.12 in. diameter with the C steels, 0.18 in. with the C-Mo steel. In a slack-quenched specimen of ordinary size, the depth to which the hard and tough structure, over 45 R.C., would be obtained, would be small, and the softer core would have the "unsatisfactory properties." Nevertheless, the tough surface might have value in a battering type of service.

Considering Elmendorf's data for 100% martensite in the light of Lyon's data for conventionally treated specimens in Fig. 234, it is seen that the tempering temperatures used by Elmendorf were not high enough to develop much ductility in the tempered martensite. The drop in properties above 50 R.C. of a 0.71 C, 0.88% Mn steel, oil-quenched in  $\frac{1}{4}$  in. diameter, and then tempered to a range of hard-

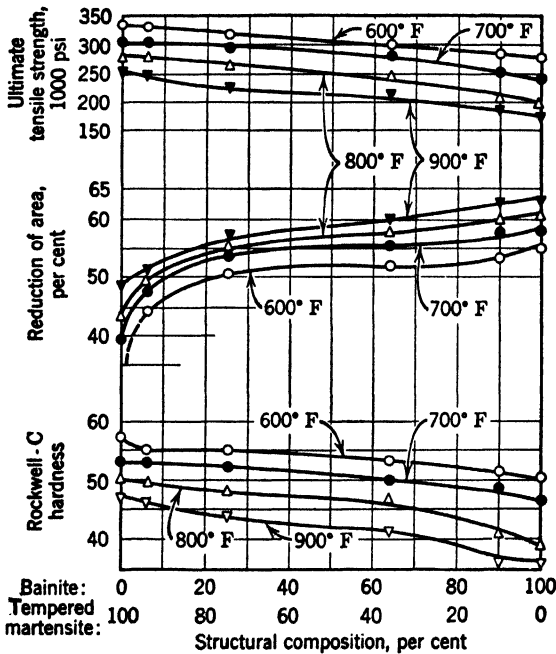


FIG. 233. Effect of structural composition on the mechanical properties of C 1080 steel. Variation regulated by controlled quenching.

ness, is shown by Lyon<sup>14</sup> in Fig. 234. On other steels the drop might come at a little lower hardness. It is with high-C content and high hardness, near, and particularly just above, this drop-off, that the austempered product shows superiority.

**Retained Austenite.** Elmendorf found that, compared to the isothermal data for transformation of all austenite, the remaining austenite in the martensite-austenite mixture produced by "time quenching" transformed to bainite more rapidly. In discussion, Klier pointed out that this was the reverse from what had been found in other steels. It was suggested that the pressure conditions set up during the transformation of austenite to martensite might be differ-

ent at different C and alloy contents. Specimen size could also make a difference. It is shown in the next chapter that knowledge of the rate of transformation of retained austenite is a confused state.

**Bainite.** The bainite produced in 1080 at 600° can hardly be termed either upper or lower bainite, since there is no sharp dividing line between the two in the S curve, Fig. 224. There is ample evidence that the

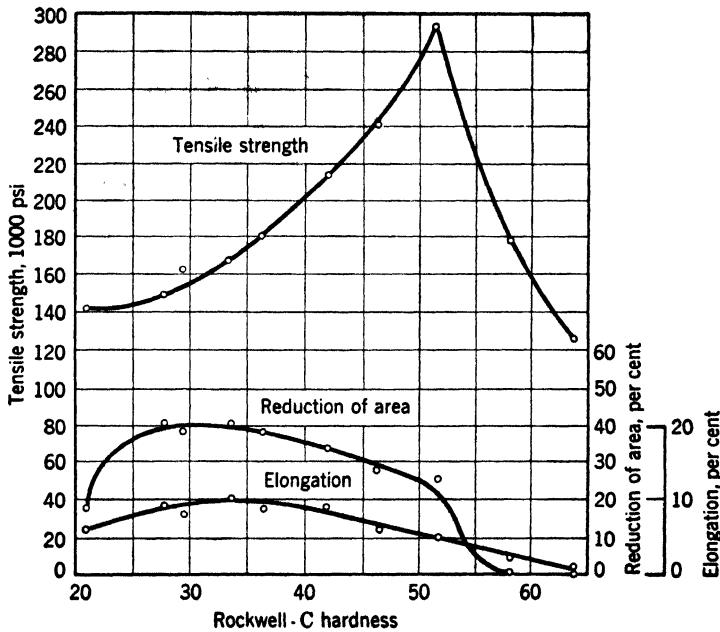


FIG. 234. Tensile strength, reduction of area, and elongation versus Rockwell-C hardness.

structure produced in a slack quench, producing true upper bainite mixed with martensite, when tempered back for times and temperatures suitable for an initial all-martensite quenched structure, softens more readily, and, at the same hardness, has lower yield strength and inferior notched-bar impact properties compared with tempered all-martensite. Most references to "bainite" in slack-quenched steels refer to upper bainite.

Like suspicion attaches to lower bainite. Yet it is not so firmly established, and it is debatable, whether acicular lower bainite, almost, if not quite, indistinguishable under the microscope from martensite, is not equivalent when tempered to martensite tempered to the same hardness. Some authorities prefer it because of the liability

for cracks to occur in rapidly formed martensite needles. "Martempering" (Chapter 19) is often carried out so as to produce more or less lower bainite. Heavy sections, whether martempered or conventionally quenched, except perhaps in the most sluggish practically air-hardening steels, are, however, likely to contain upper bainite. In heavy armor, where proper structure is important in the middle of the plate as well as at the surfaces, walking the chalk line between definitely harmful upper bainite and desirable or at least not very harmful lower bainite, mixed with martensite, is ticklish business, complicated by the fact that pressure conditions are involved in these large sections, as is shown by the failure of like cooling rates to produce the same structures in Jominy bars and in heavy armor.

Austempering deserves to be kept in mind as a possible solution for specific heat-treating problems, but its applicability is limited to small stock, to be used at high hardness. Hollomon and coworkers<sup>15</sup> utilized isothermal treatment to produce a variety of structures corresponding to different degrees of slack quenching, which were then variously tempered and subjected to the conventional mechanical tests. This set of experiments provides an illuminating easily grasped confirmation of the previous mass of scattered evidence on the virtue of avoidance of slack quenching.

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## CHAPTER 19

### QUENCHING, MARTEMPERING, $M_s$ - $M_f$ , RETAINED AUSTENITE

The various factors affecting the stabilization of austenite are given in schematic form in Fig. 235, summarizing the discussion in previous chapters. Once austenite has been retained down to the

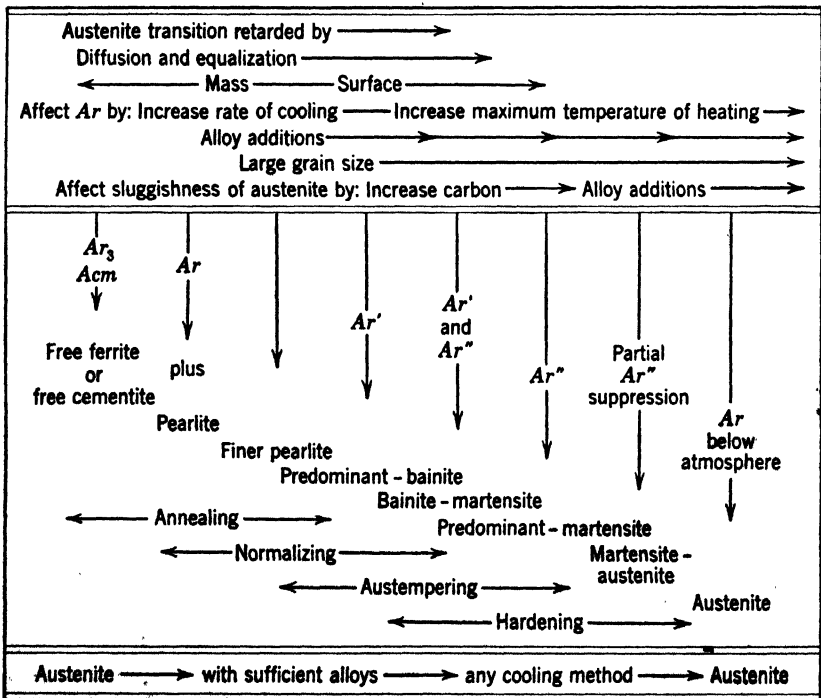


Fig. 235. Factors affecting the transformation of austenite.

temperature of martensite formation,  $M_s$ , and cooling is thereafter continued, the formation of martensite proceeds at an extremely rapid rate. No time delay intervenes as it does in the transformation to ferrite, pearlite, and bainite. The operation of quenching is aimed to secure a fast enough cooling rate to retain austenite unchanged

down to  $M_s$ ; then, if cooling is continued through  $M_f$ , nature does the rest. When the section is so large that it is impossible to cool the center at the necessary "critical rate" to retain austenite in plain C steels, alloying must be resorted to. The presence of most alloys, however, lowers  $M_s$ . Martensite is less dense than austenite, and the volume change introduces internal stress. Untransformed austenite presumably cushions the stress. Gradual dropping of the temperature through the range  $M_s$  to  $M_f$ , from the start to the finish

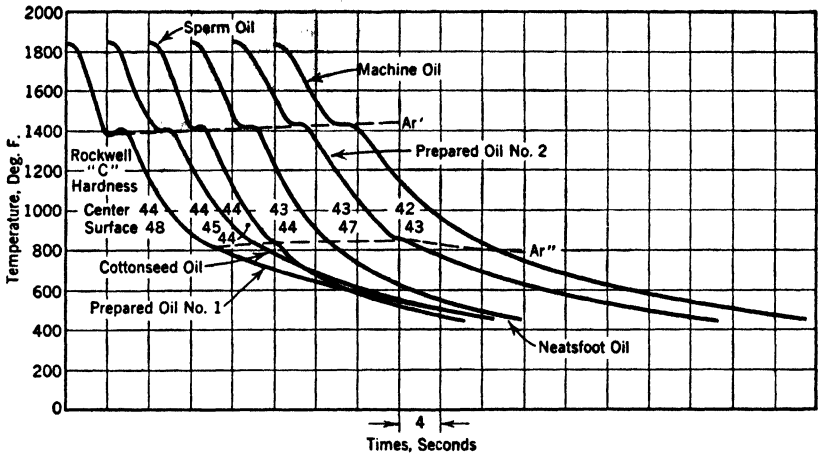


FIG. 236. Time-temperature cooling curves at center of  $\frac{1}{2}$ -in.-diameter cylinders of 0.95% C steel, quenched from 1605° into various oils motionless at 70°. (French and Klopsch)

of production of martensite, is presumed to allow continual adjustment by pushing the relatively plastic austenite around without producing localized overstress that ruptures the martensite. Rupture of individual martensite needles has been noted, and bodily rupture of the steel, either during quenching or soon thereafter, is an all-too-common occurrence.

Cold-water quenching gives relatively rapid cooling from  $M_s$  down. Hot water has a much too slow cooling rate when coming down past the pearlite nose of the true S curve, but it cools fast within the  $M_s$ - $M_f$  range. Oil cools regrettably slowly past the pearlite nose, but acceptably slowly through  $M_s$ - $M_f$  (see Figs. 6 and 236).

Hence, when water quenching produces cracks, the heat treater turns to oil quenching, and chooses an alloy steel, sluggish enough to stand the slow cooling at the pearlite nose, in order to make oil quenching applicable. More detailed discussion of quenching media,



pressure quenching, and quenching cracks is deferred to Vol. II, and it is merely noted here that heat can be abstracted not only by these usual coolants, but also by gases and solids. So a gas stream, a cold metal die forced into contact with the piece to be quenched, or, in welding, the parent metal next the heat-affected zone can be used to drain heat away and, if the draining is rapid enough, may thus act to produce martensitic quenching.

We need not linger long in discussion of quenched martensite itself, for the use of untempered martensite is limited to that of a wear-resistant surface, as on a carburized piece, and even then it is ordinarily given at least a stress-relief tempering. Carburizing and heat treatment of carburized parts are discussed in Vol. II.

Martensite is intensely hard and intensely brittle, and so, for most engineering uses, it is tempered to provide toughness.

### MARTEMPERING

The hypothesis that there are internal cracks in martensite or internal stresses in it, either case supposedly resulting from the large and rapid dimensional change that occurs when martensite forms in quenching, plus the fact that actual cracking of the piece often occurs during or just after a severe quench, leads to speculation as to possible benefits from having a less sharp temperature gradient within the piece when it reaches the temperature of martensite formation.

**Cooling Rates.** The difference between the surface and center-cooling rates in conventional quenching has been brought out in Figs. 184-186. To help the center to lose heat more rapidly during the actual quench it is logical first to give it time for coming down to the temperature of the outside, thus flattening the temperature gradient, while the steel is still austenitic. Obviously, the best holding temperature will be down in the bay of the true S curve (that for cooling conditions) just above the temperature for rapid formation of bainite, and will vary with different steels. If the steel is such that the center cannot be cooled rapidly enough to retain austenite with a vigorous quench, so that the desired low holding temperature cannot be used, some benefit can be had by holding at a temperature above the pearlite nose.

**Temperature Equalization.** Angell<sup>1</sup> has studied this matter on a wide variety of steels, and in no case were the properties injured (provided, of course, that the piece was first fully austenitized) by quenching from the lower temperature, whereas cracking was greatly diminished. For example, a 0.23 C, 4.61% Ni steel was quenched

(a) from 1450°, and (b) from 1150° after having been first held at 1450°. Both lots were then tempered at 900°, the tensile, impact, and hardness results being essentially identical, in spite of the 300° difference in quenching temperature. On some highly alloyed steels it is even possible to quench from a temperature lower than that used for tempering after the quench.

The holding can either be in a furnace at the selected temperature, or in a fused bath. The piece can be "quenched" into the fused bath, though this quenching is not to harden, but merely to secure temperature equalization through the section. The hardening is produced by cooling, either in air or in another quenching bath, after removal from the fused bath.

**Early Experiments.** Quenching of 0.192-in.-diameter wire of 0.79 C, 0.45% Mn in fused salt at 450° was studied by Lewis<sup>2</sup> back in

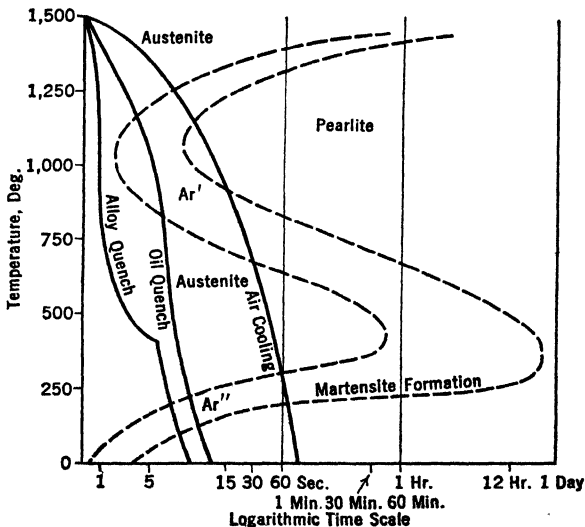


FIG. 237. Air cooling, oil quenching and molten bath quenching of small-diameter spring wire, compared with the isothermal S curve. The martensite lines are incorrectly located. (After Trautman)

1929. At this temperature (in this size), austenite is retained to be transformed to martensite on air or oil cooling. The method was suggested by Lewis for avoidance of quenching cracks and distortion.

Trautman<sup>3</sup> discussed the heat treatment of 0.60–0.70 C, 0.90% Mn spring wire in 0.091 in. diameter by passing an electric current through it, thus heating it uniformly through its cross section, to 1450°, the operation requiring only 15 sec; then quenching into a

molten low-melting alloy bath held at 425° for about 4 sec, and thence to oil where the cooling is completed.

Hardness of 64–66 Rockwell C is obtained. The hardness is reduced to 45–47 Rockwell C by tempering at 800°.

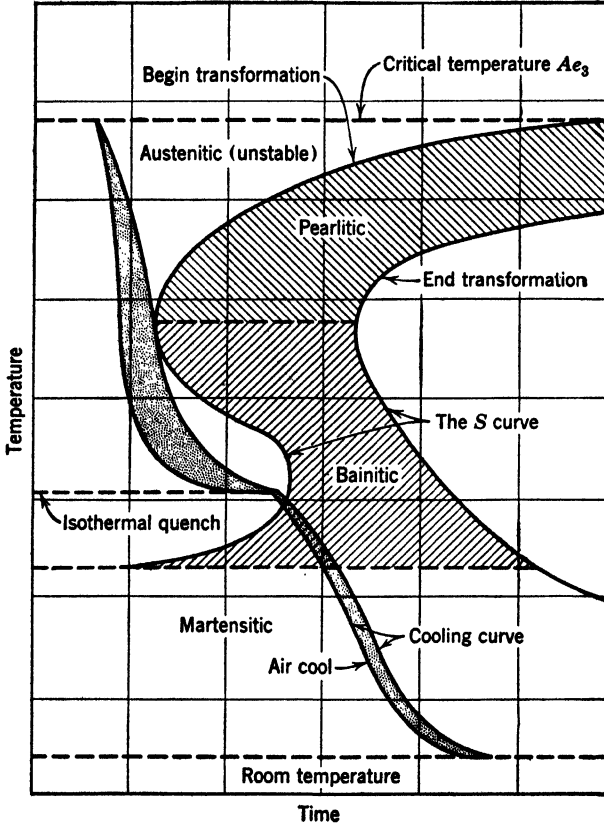


FIG. 238. The cooling curve in the quench and temper method in relation to the S curve.

Comparative data on 0.091-in.-diameter wire of 0.63 C, 1.01 Mn, 0.19% Si, after quenching and tempering to Rockwell C 47–48 by this method and the usual oil-quenching method, are cited in Table 42.

TABLE 42

Quench	Tensile	Johnson Elastic Limit	Elong., %		R.A., %	Torsion Twists in 10 In.
			10 In.	2 In.		
Oil only	248,000	203,000	3.4	6.3	53	16
Alloy, then oil	248,000	229,000	3.1	5.2	55	25

It is stated that, owing to the very rapid rate of heating in the electric method, high quenching temperatures, even  $1650^{\circ}$ , may be used without grain growth. Contrasted with heating the wire by pulling it through a lead bath or through a heated enclosure, the

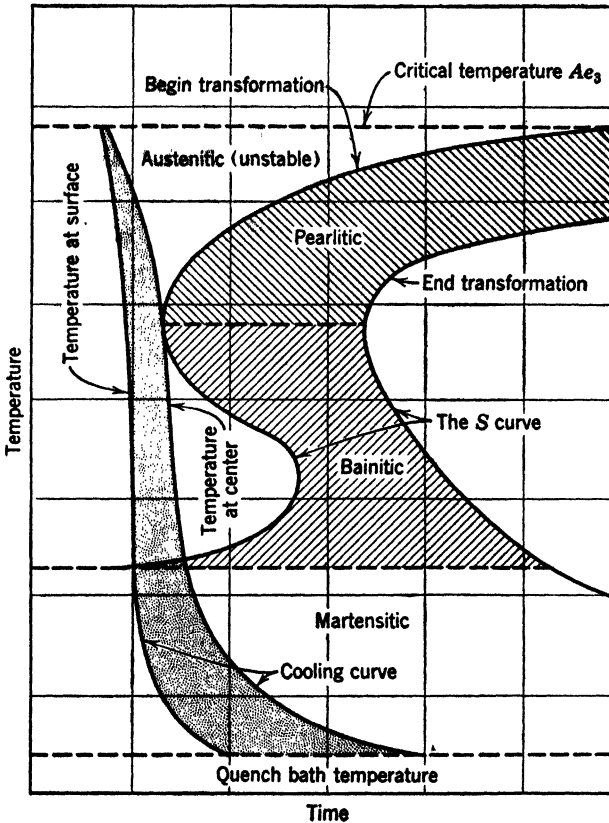


FIG. 239. The cooling curve in the martempering method in relation to the S curve.

direct electrical-resistance method heats the whole cross section uniformly instead of the heat having to soak in from the outside. It may be noted that the cross section of the wire treated by alloy quenching was small.

Although the proper plotting of the martensite lines of the S curve was not then understood, Trautman's diagram, Fig. 237, summarizes the process of temperature equalization before  $M_s$  is reached, a process later given the term "martempering," a misnomer, for the martensite formation comes only in the final cooling, and there is no tem-

pering involved; the regular tempering treatment is applied after martensite has been allowed to form.

With a steel and a section size such that the pearlite nose is missed at the center of the section, one could quench into a fused-salt or in a hot-oil bath. At a temperature in the "bay" where the time delay for bainite formation is long enough for reasonable temperature equalization, and the austenite remains wholly untransformed, it becomes ready for changing to martensite by air cooling. Or, by leaving the piece in the hot bath, it could be transformed to bainite. It is this choice that probably led to the term, "martempering," when cooling is to be done, since "austempering" is the term used when the piece is left in the bath to transform to bainite.

The principle involved, the fact that it is the center-cooling curve rather than that at the surface, and the chief drawback of "martempering," that is, its tendency to produce some bainite instead of all-martensite, are all brought out by Babcock<sup>4</sup> in Figs. 238-239.

Important points are how long it takes to equalize the temperature in the piece and how successful a fused-salt bath will be in forcing cooling past the pearlite nose.

Shepherd<sup>5</sup> gives the time in minutes for equalization of center and surface temperatures of rounds quenched from 1550° into a circulated fused-salt bath held at the temperatures stated as shown in Table 43.

TABLE 43

<i>Salt Temperature</i>	400°	500°	600°
1-in. round	5	4	3½
2-in. round	8	7	6
3-in. round	13½	12½	11½

**Martempering in Various Coolants.** Shepherd shows surface and center hardness (Fig. 240) of cylinders of different sizes of NE 8744 quenched in various coolants, ranging from brine at room temperature to fused salt at 400°. These show that both the surface and center hardnesses on quenching into fused salt are much below those produced by conventional quenching into water or oil at room temperature, but higher than quenching into oil held at 400°. A variant is to quench into hot oil. This retards the cooling rate within the  $M_s$ - $M_f$  range, so as to produce martensite more slowly, at the same time producing some lower bainite. This sacrifices some hardness but tends to minimize distortion. The cooling rate depends on the

section, but, to show the order of magnitude, Miner<sup>6</sup> finds that cooling a 7/8-in. round from 1550° to 400° takes about 36 sec in regular quenching oil at room temperature; in an oil compounded for stabil-

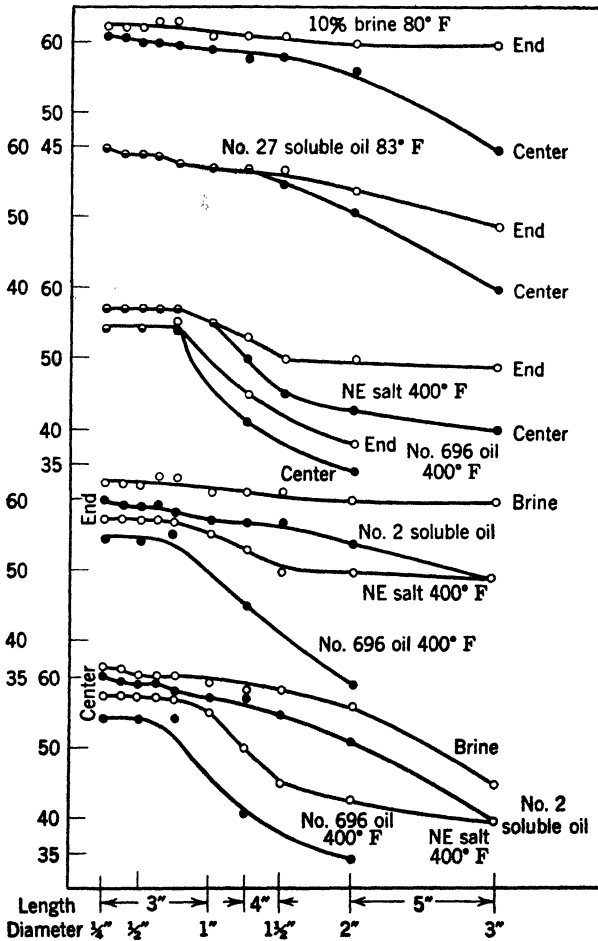


FIG. 240. Transverse hardness of various-sized rounds of N.E. 8744 steel when quenched from 1500° in brine, soluble oil, salt and oil at 400°. The N.E. 8744 analysis is 0.43 C, 0.87 Mn, 0.03 P, 0.008 S, 0.50 Cr, 0.46 Ni, and 0.27% Mo.

ity at such temperature and held at 300°, cooling takes a little over a minute. This gives opportunity for production of some bainite. Miner cites a 1 7/8-in.-diameter shaft of 1340, which fully martensitic, would show 61 R.C. Quenched into oil at 300°, it had 56 R.C. at the surface, 53 R.C. at the center. If it were completely transformed to

lower bainite just above  $M_s$ , it would show around 45 R.C. The structure, before tempering, is therefore a martensite-bainite mixture with a good deal of bainite at the center.

The utility of such a treatment depends on the relative importance of the diminished warpage and the lowered yield strength.

The relative proportion of martensite and bainite produced by retarded cooling through  $M_s$ - $M_f$  will vary greatly with different steels, according to their  $M_s$  and  $M_f$  temperatures and their S curves, as well as with section size. Austenite retention also needs to be considered.

Brown<sup>24</sup> cites a salt quench for temperature equalization in carburized pieces, followed by air cooling for hardening the case while leaving the core soft for machining and minimizing distortion. For pinions of 4620 the salt quench after gas carburizing was at 400°; for a gear of 3312 it was at 1000°. Thorough stress-relief anneals were given prior to carburizing.

**Time Quenching.** Another theoretically possible way to quench rapidly past the pearlite nose in water or brine, and yet avoid too rapid cooling in the  $M_s$ - $M_f$  range, is to use a layer of oil on top of the aqueous layer, plunge the piece into the lower layer, and at just the proper split second pull it up into the oil layer, or to transfer the piece from the water quench to a separate oil quench. Obviously, this is an extremely tricky procedure (compare p. 477).

**Limitations.** The bug in the application of these principles lies in the necessity, when any but very small sections are treated, for a rather highly alloyed sluggish steel in order that at the relatively slow cooling rate of the hot-salt or hot-oil quench, the pearlite nose of the S curve will be missed, and also that upper bainite will be avoided. Thus, in such processes, the less hardenable cheap, C and low-alloy steels have size limitations somewhat similar to those for austempering.

In martempering it is imperative that the  $M_s$  temperature be known, for it is desired to produce temperature equalization above it. A knowledge of the bainite portion of the S curve is also needed. If the time delay in the bay, between the pearlite nose and the bainite chin is small, bainite will form before the desired temperature equalization can be accomplished. Hence, any section too large for its center to cool to the necessary temperature within the time delay, set by the sluggishness of the steel, is unsuited to martempering, if bainite is to be avoided.

$M_s$ —START OF THE MARTENSITE TRANSFORMATION

Other things being equal, the higher the  $M_s$  temperature the better, on account of the propensity toward cracking when martensite forms at low temperature. If it is desired, as it usually is, to transform all the austenite to martensite by a conventional quench,  $M_f$  should be above room temperature. As is discussed in Vol. III, Mn, Ni, Cr,

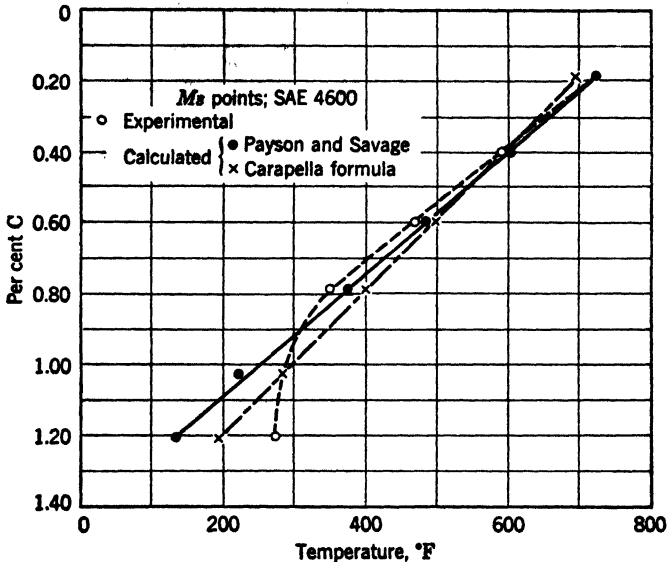


FIG. 241. Martensite points for 4600, determined experimentally and calculated from chemical analysis.

and Mo materially lower  $M_s$ , whereas Si does not affect it, and Cu has very little effect.

Payson and Savage,<sup>7</sup> Carapella,<sup>8</sup> and Rowland and Lyle<sup>9</sup> presented formulae for calculation of the  $M_s$  temperature of an alloy steel. As Fig. 241 shows, the formulae did not give correct results on the 4600 series.

Shepherd<sup>5</sup> assembled calculated  $M_s$  points for various C contents in various alloy steels, although the calculations for hypereutectoid steel would be acceptable only for complete austenitization.

Payson and Savage point out that the formulae are "useful only to approximate the values and not to measure them."

Grange and Stewart<sup>10</sup> use a slightly different formula for estimating  $M_s$  from those proposed by Payson and Savage, Carapella, or



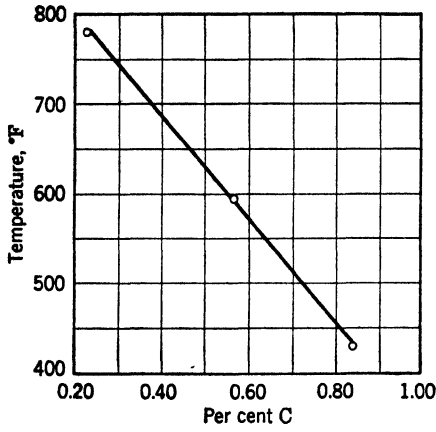


FIG. 242. The effect of C on the  $M_s$  of plain C steel.<sup>6</sup>

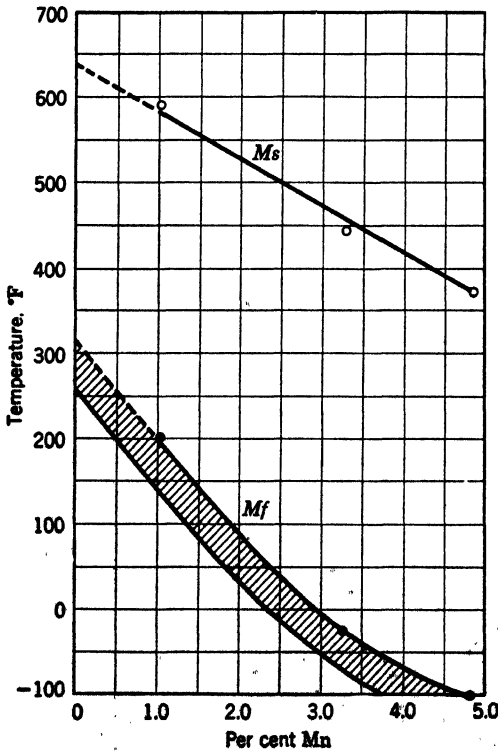


FIG. 243. The effect of Mn on the  $M_s$  and  $M_f$  of 0.50% C steel.

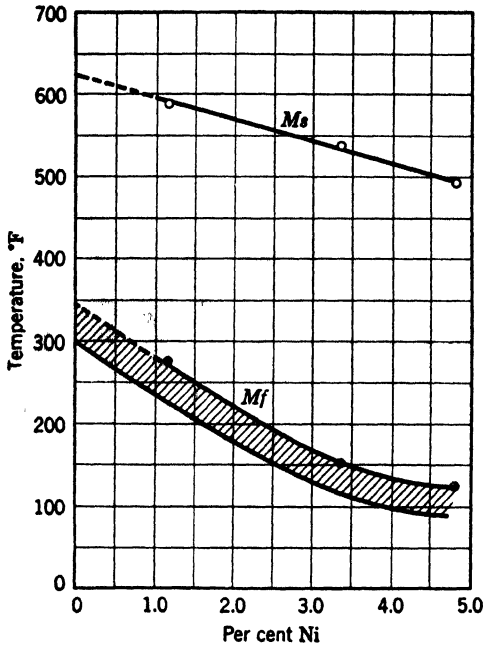


FIG. 244. The effect of Ni on the *M<sub>s</sub>* and *M<sub>f</sub>* of 0.50% C steel.

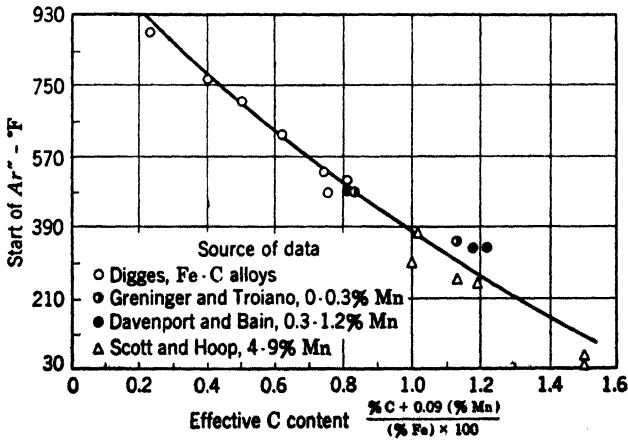


FIG. 245. Start of *A<sub>r</sub>''* (*M<sub>s</sub>*) in Fe-Mn-C alloys so cooled as to retain all C in solid solution.

Rowland and Lyle. For NiCrMo steels, providing all carbides have been fully dissolved in austenitizing, and provided that C is between 0.20 and 0.85%, Mo below 1%, and Cr not higher than 1½%, they suggest  $M_s + (0.025 \times \% C) - (70 \times \% Mn) - (36 \times \% Ni) - (70 \times \% Cr) - (50 \times \% Mo)$  as giving  $M_s$  in degrees Fahrenheit

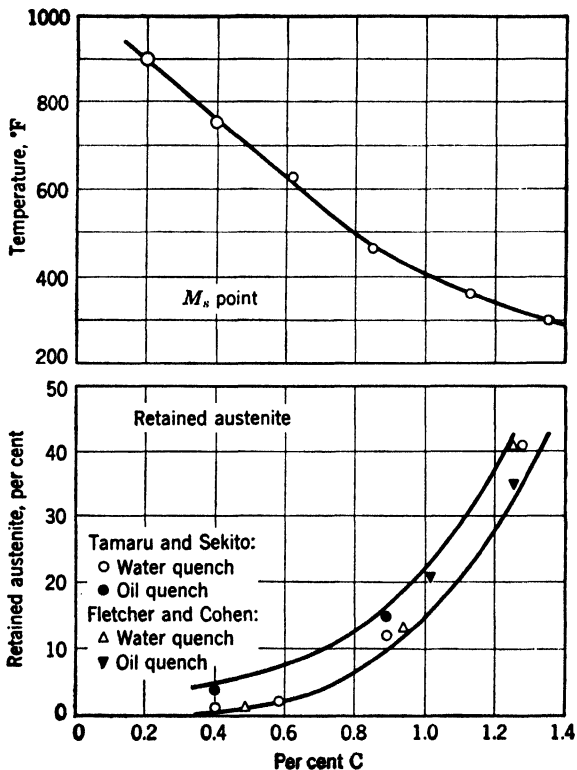


FIG. 246. Effect of C content on the  $M_s$  point<sup>21</sup> and the retained austenite content.<sup>11</sup>

heit. For selection of a martempering temperature, such a calculation will presumably be adequate; yet, even experimentally,  $M_s$  is not yet determinable with great accuracy, for the dilatometer data of Christenson and coworkers<sup>22</sup> do not agree very well with metallographic data. It may incidentally be noted that some English writers denote the "martensite arrest" by  $A_m$ , instead of  $M_s$ .

Carbon is the most potent element in reducing  $M_s$ , which accounts for the practice of introducing alloys into medium-C steels to get hardenability, instead of raising C.

The effect of C alone is shown in Figs. 242 and 246 (which do not agree exactly), and that of Mn and Ni alone in Figs. 243 and 244. The combined effect of C and Mn is shown in Fig. 245. Figure 246 indicates that, as *M<sub>s</sub>* falls, retained austenite rises.

Grange and Stewart studied the *M<sub>s</sub>*–*M<sub>f</sub>* range by the expedient of immersing a small austenitized specimen in a fused metal bath, held

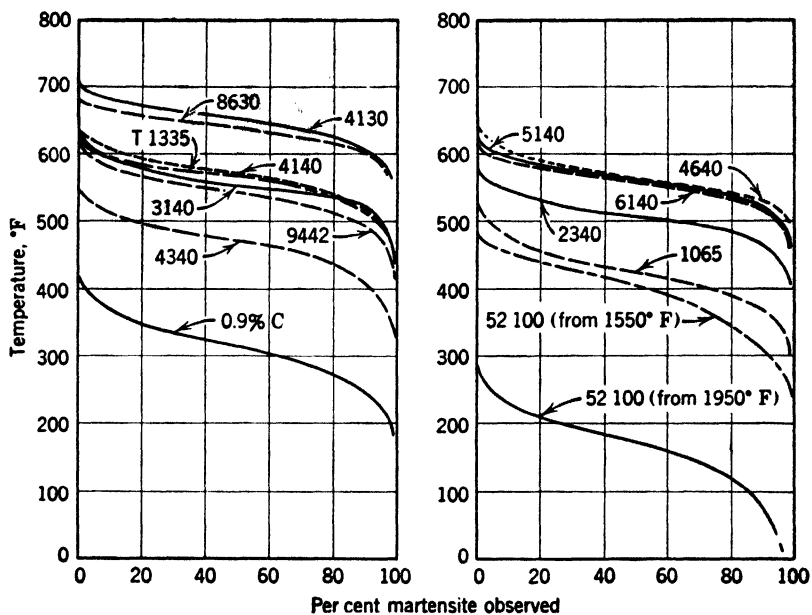


FIG. 247. Relation of percentage of martensite to temperature in 14 commercial grades of steel.

at a definite temperature in the range from a little above the expected *M<sub>s</sub>* temperature, to one below the expected *M<sub>f</sub>* temperature, producing the amount of martensite characteristic of that temperature then tempering it for 5 sec at 1200° (without having previously lowered the temperature), then brine-quenching to transform the residual austenite to martensite. The specimen is picral-etched. The original martensite, after tempering, appears dark, and the second lot of martensite, white, under the microscope.

Results for 14 common steels are shown in Fig. 247. These curves tail off sharply at the right; that is, *M<sub>f</sub>* is not easily determined.

Once *M<sub>s</sub>* has been estimated by calculation or determined directly, it would be nice to be able to estimate *M<sub>f</sub>* by subtracting the proper

number of degrees, if we knew what this number was, but there is much uncertainty about it.

Grange and Stewart demonstrate this by the scatter in Fig. 248. The shaded bands in Figs. 243 and 244 also indicate the scatter.

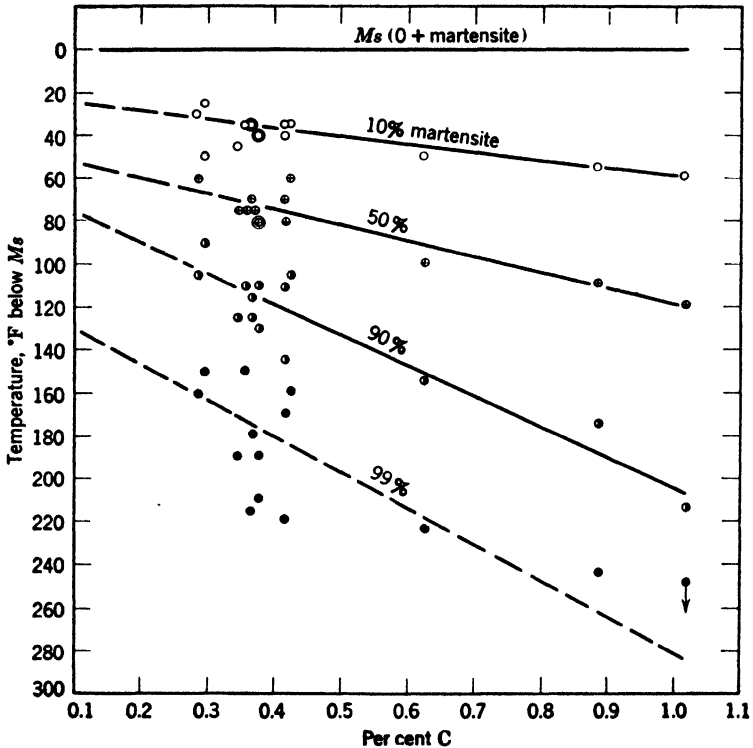


FIG. 248. Chart for estimating proportion of martensite at any given temperature when  $M_s$  and C content are known.

The variously alloyed 0.40% C steels that were studied have  $M_f$  anywhere from 150 to 220° below  $M_s$ . In common with other observers, Grange and Stewart find that austenite grain size does not affect the  $M_s$  temperature, nor the proportion of austenite transformed to martensite at any temperature between  $M_s$  and  $M_f$ . However, although the proportion remains the same, the martensite needles coming from coarse austenite are larger than those from fine austenite. Grange and Stewart comment that quench cracking is likely to be worse with coarse austenite.

There are limitations and uncertainties connected with this method of determining  $M_s$  and  $M_f$ , because, as is also the case in producing

conventional S curves, tiny specimens are used, and so the pressure conditions met in large specimens are not duplicated. As the authors point out, 3140, for example, retains much more austenite at the center of a large quenched piece than would be predicted by Fig. 247.

Eddy, Marcotte and Smith<sup>11</sup> reported that  $M_s$  temperatures ran from 50 to 100° lower for the cast than for the wrought condition. This would hardly be expected and needs substantiation. They determined the S curves for the same heats of 1030, 2330, 4130, and 4330 in both cast and wrought conditions, finding close agreement. Jominy curves likewise showed close agreement.

Though, in most steels,  $M_s$  is not influenced by grain size or quenching temperature, in hypereutectoid and especially in high-speed steel,  $M_s$  and  $M_f$  are both lowered as the austenitizing temperature (hence the degree of carbide solution and homogenization) is increased.<sup>20</sup>

#### $M_f$ —COMPLETION OF THE MARTENSITE TRANSFORMATION

The finishing of the martensite transformation, the destruction of *all* the austenite, at  $M_f$ , is not so clear cut a matter, but it is an important one, since, if the quenching is not continued clear down to  $M_f$ , austenite is retained. The retained austenite, heated above  $M_s$  in tempering the originally formed martensite, is ready to make a fresh start and, in its turn, transforms to brittle martensite on cooling from the tempering temperature, so that a second tempering is needed to soften these second-stage martensite patches.

Or, if we seek to transform all the austenite by "cold treatment," prior to tempering, we need to know the  $M_f$  temperature so as to select adequate means of refrigeration. Payson and Savage find that the  $M_f$  values are uncertain, but approximately 325 to 475° below  $M_s$ , with the spread between  $M_s$  and  $M_f$  the greater for the elements most effective in depressing  $M_f$ , that is, Mn, Cr, and Ni.

**S Curves in Relation to Martempering.** Since the completion of the martensite transformation requires a falling temperature and is not accomplished by isothermal holding, while the transformation to bainite does occur at a constant holding temperature, in a steel where the bainite reaction can occur in the temperature range of  $M_s$  to  $M_f$ , the cooling conditions will govern the extent to which each transformation occurs. Water quenching will favor martensite formation, and oil quenching that of bainite, because of the shape of the cooling curves in water and oil in this temperature range.

If the lower (bainite) portion of the S curve swings toward the left in the  $M_s$ - $M_f$  temperature range, the bainite reaction will get in the way of the martensite reaction and use up austenite. The bainite reaction has a strategic advantage over the martensite reaction in that it keeps going on at any pause or slowing up of the cooling, whereas martensite formation requires a decrease in temperature; it ceases when cooling ceases, and slows up when cooling slows up.

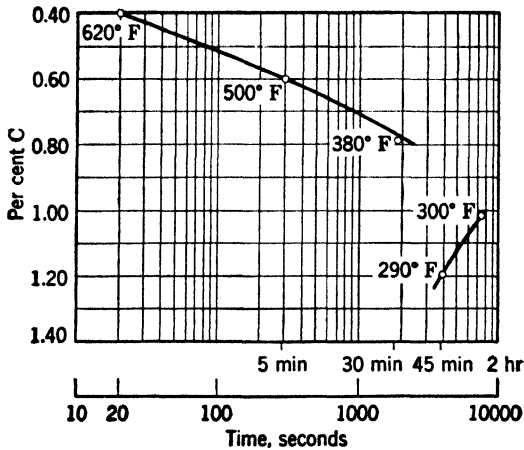


Fig. 249. Maximum holding times for 4600 in quenching baths held about 20°F above  $M_s$  points of respective C contents to avoid production of bainite.

Hence, the disturbing factor that leads to slack quenching and to variations in quenching behavior among alloy steels is best brought to light by examining the lower portions of their S curves, whenever these are clearly established, which is all too infrequently. This complication is recognized by Rowland and Lyle,<sup>9</sup> who were studying the applicability of the determination of  $M_f$  for showing the case depths and C gradient in a carburized 4620 steel. (This applicability, as well as that to the reverse case of decarburization, had been pointed out by Payson and Savage.<sup>7</sup>) The technique was to hold at a temperature about 20° above the  $M_s$  point (as would be done in martempering), and then quench into hot baths at various lower temperatures. If the austenite were held longer than the times shown in Fig. 249, bainite began to form. That is, 4640 starts to form martensite at 600°, but, if allowed to remain at 620° for more than 20 sec, the austenite starts to change to bainite.

## RESIDUAL AUSTENITE AND ITS DECOMPOSITION

Discussion of  $M_f$  brings us up against the problem of austenite retained on quenching, and what happens to that retained austenite if the temperature is again raised above  $M_s$  and a second cooling occurs. Knowledge is scanty and understanding even more so.

It is logical to expect that, since sufficiently fast cooling can retain austenite down to the temperatures of bainite and martensite formation, since completion of the bainite reaction is slow in many alloy steels, and since martensite forms on a falling temperature, some steels, in normal heat treatment, will retain some austenite at room temperature. This is true, and in carburized parts the case is especially prone to retain austenite (see Fig. 246).

A peculiarity about austenite retention, long ago noted, especially by Mathews,<sup>31</sup> is that, rather unexpectedly, oil quenching tends to retain more austenite than water quenching. The explanation probably lies in the slower cooling rate in the low-temperature ranges, those of bainite and martensite formation, in oil than in water. Although this longer opportunity for the transformations to go on might be expected to lead to more complete formation of martensite, this is not the case; for it must be remembered that martensite forms only on a falling temperature.

**The Pressure Effect.** With more time at any temperature level within the martensite-formation range of temperatures, there is more opportunity for stabilization of patches of retained austenite due to pressure from the increase in volume of the surrounding martensite. With a more rapid rate of cooling through the martensite-formation range, and quicker jumps down each step in the cooling, whatever austenite is retained is less stabilized. There is less balancing of pressure from all sides. Moreover, the contraction on cooling tends to relieve the pressure, which may account for the necessity for a falling temperature.

The pressure effects are not independent of section size; Scott<sup>12</sup> noted that cracking of quenched 1% C tool- or alloy-steel cylinders, ascribed to internal stress, was greater in water than in oil up to  $\frac{1}{2}$  in. diameter, but both coolants produced cracking in  $\frac{3}{4}$  in. and the 1 in.-diameter-cylinders cracked in oil but not in water. Slack-quenched larger cylinders did not crack in either. Tensional stress at the outside accompanied cracking; compressional, freedom from it.



Hence, the information<sup>13</sup> of Fig. 250 cannot be generalized to cover other sizes than the one used.

Tied up with the topic of retained austenite is that of the condition of the martensite. Microcracks may appear in martensite needles<sup>14, 15</sup> and may not heal up on tempering. Some pools of retained austenite may act as cushions to minimize the stress on the martensite needles

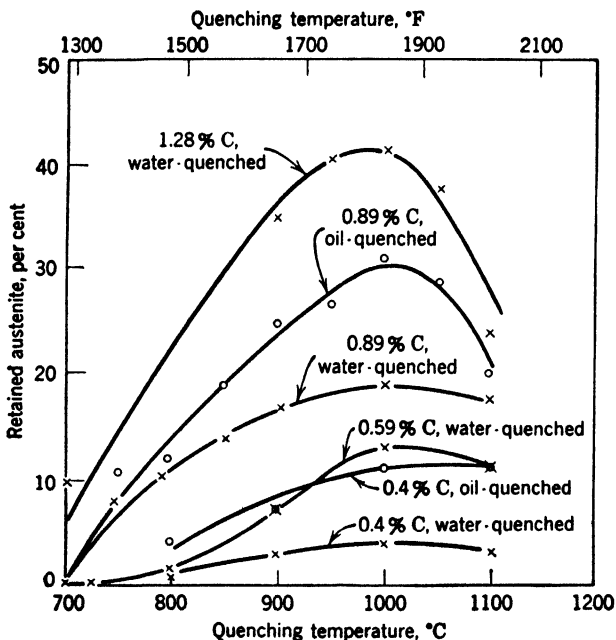


FIG. 250. Amount of retained austenite in quenched cylinders 5 mm in diameter, as determined by X-ray measurements. (Tamaru and Sekito)

themselves and decrease the propensity toward development of these internal cracks. But Gurry<sup>16</sup> shows an excellent micrograph of a 1.39% C FeC alloy, brine-quenched from 1750° in which some retained austenite has not prevented the cracks.

**Stress Effects.** The way the still-existing austenite behaves seems to depend on stress. An air-hardening 0.50 C 5% Ni steel readily retains austenite, as Fig. 251 shows from the hardness figures. This plot is for a small section. Gardner and coworkers<sup>17</sup> show that the soft constituent in such steels is really austenite and that in the order, air cooling, oil quenching, water quenching, more austenite is retained by the slower cooling. On tempering at 500°, the retained austenite is slowly transformed to bainite. It is under different stress

conditions during heating and no longer refuses to transform. Of course, the martensite is simultaneously tempering. As much as 5 or 10 per cent of austenite may be retained to exist along with the martensite formed in an ordinary quench with the coolant at room temperature, in some everyday heat-treatable steels. In these amounts, the austenite is so distributed that its presence does not drop the hardness readings; for all the hardness tells us, we would think we had all-martensite.

Thus a structure which was martensite plus austenite on quenching becomes tempered martensite plus bainite on long tempering.

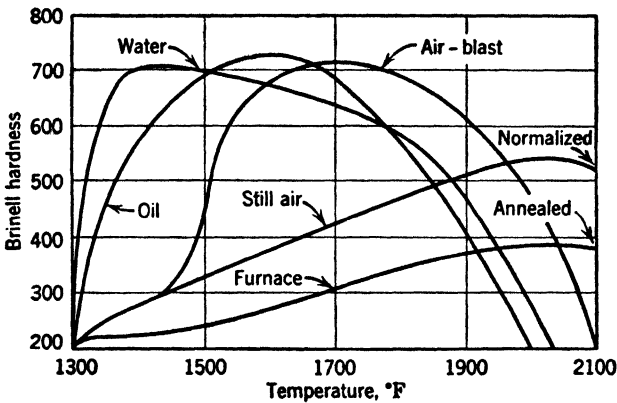


FIG. 251. Hardness of 0.50 C, 5% Ni steel quenched or cooled in various media from the temperatures shown. (After Jungbluth<sup>33</sup>)

The bainite in turn will temper. Prior to the tempering of bainite, there is an initial hardening due to tetragonal martensite shifting to cubic martensite or to other effects coincident therewith, then a tempering and softening of the cubic martensite, then the soft residual austenite goes over to the harder bainite, with a second hump in the hardness curve (Figs. 252<sup>18</sup> and 253<sup>19</sup>); then all the constituents temper.

If tempering is stopped before all the residual austenite has changed to bainite, and the steel is cooled, the "residual of residual" austenite should now be able to transform to martensite, since the tempering has released much of the internal stress. In high-speed steel, it is supposed to, though, according to Antia and Cohen,<sup>20</sup> it does not in all ordinary steels. On the other hand, Elmendorf<sup>21</sup> warns that, in such a case, sufficient transformation time must be allowed for decomposition of retained austenite into bainite, lest the

retained austenite, after cooling, be present as untempered martensite. In discussion, Klier<sup>22</sup> points out that "austenite in mixed structures of various sorts transforms at a lower rate than 100% austenite."

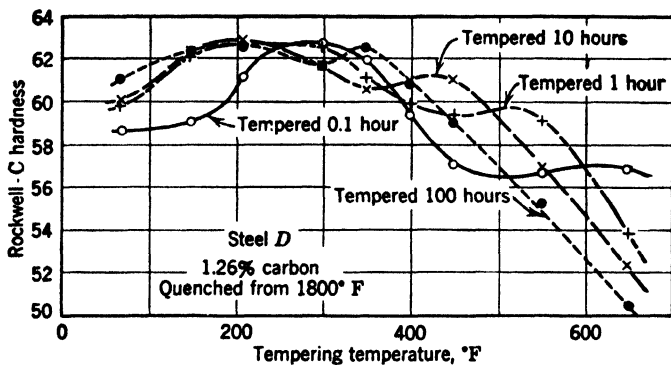


Fig. 252. Hardness-tempering curves for 1.26% C steel, 35% retained austenite.

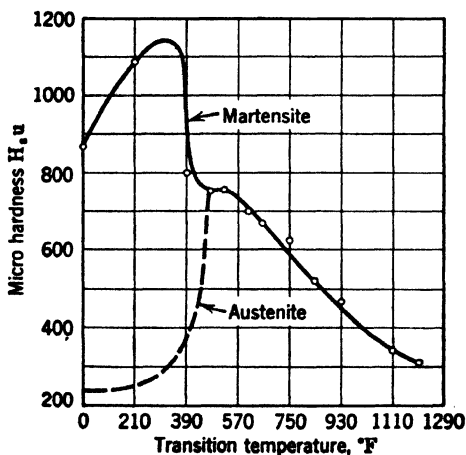


Fig. 253. Microhardness of martensite, of austenite and of their transition phases in a steel of 1.7% C content, depending on the temperature reached in the tempering process. Time of heat treatment 1 or  $\frac{1}{2}$  hr.

**Effect of Carbon.** The increasing tendency for austenite retention with increase in C is nicely shown by Scott's<sup>28</sup> curve, Fig. 254, for carburized 3312, oil quenched, and tested for hardness on a taper-ground specimen before and after cooling in dry ice.

The production of internal stress on a microscale within a layer of martensite enclosing some untransformed austenite, its production on

a macroscale in a slack-quenched steel with a softer center, and the modification of, or even the reversal of, the sign of the stresses on heating, with its resultant thermal expansion, and the change in volume due to transformation of martensite to products occupying less volume, and of residual austenite to products of different volume, make the situation complicated and obscure.

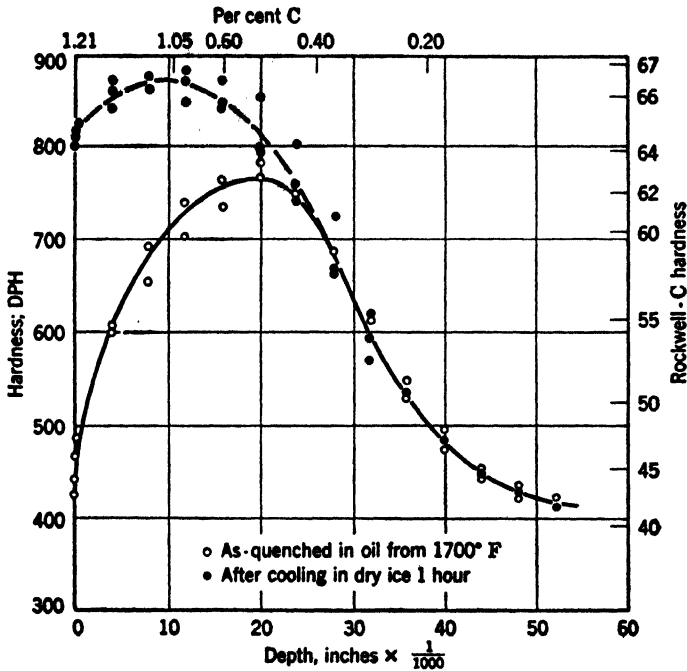


FIG. 254. Hardness variation with depth below surface in 1-in. cube of 3312 steel gas-carburized 6 hr at 1700°F and quenched directly in oil.

Considerable differences in magnitude and distribution of internal stress occur, even without the added complication of martensite formation and its change in volume, as Fig. 255, from Johnston,<sup>24</sup> for rails cooled at different rates demonstrates.

Patches of retained austenite that will be stable under the conditions of service may be useful, as French<sup>25</sup> points out, but patches of retained austenite which can transform to untempered brittle martensite would often be deleterious. Severe surface cold working of a completely austenitic structure, such as 18:8 or Hadfield Mn steel, produces a thin hard layer, usually considered to be martensite, with wear resistance resulting from its hardness, but Caille<sup>26</sup> reported

that, in the case of pins in chain links, fatigue cracks may develop in this brittle surface. Compare p. 167. The surface temperature might be raised above the  $M_s$  point during "cold" working, and martensite might be formed on cooling. Goss,<sup>27</sup> however, produced evidence that the surface layer is more likely just to be cold-worked austenite, a conclusion with which the discussers did not all agree.

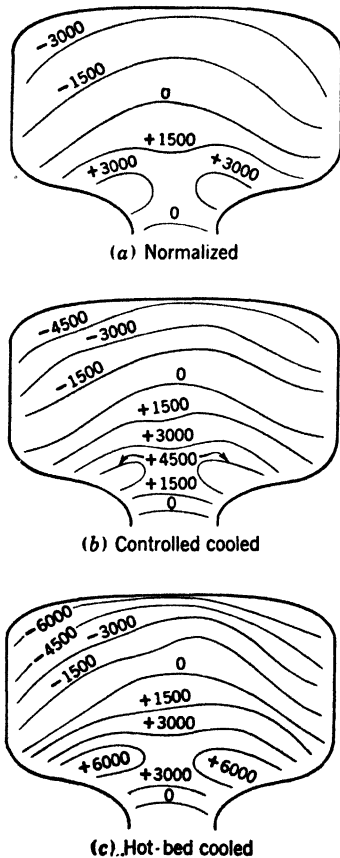


FIG. 255. Contour map of measured residual stresses, in pounds per square inch, in head of rail which has been (a) normalized, (b) cooled slowly, (c) cooled on hot bed. Tensile stress, +; compressive stress, -.

The temperatures of the  $M_s$  and  $M_f$  points obviously influence the opportunity for completion of the martensite transformation during cooling to room temperature. Less austenite is retained when the  $M_f$  point is high, as in medium-C low-alloy steels than when it is low, as in high-C high alloy steels. Figure 243 indicates that in a 0.50 C, 1% Mn steel cooling to 150°F would transform all the austenite, though the band for  $M_f$  indicates uncertainty, and Fig. 246 indicates that, with water quenching very little is left. The trend of the curve in Fig. 246 shows that at 0.70% C, something of the order of 3% with water, 10% with oil, would be retained. Fletcher and Cohen,<sup>18</sup> working with  $\frac{3}{8}$ -in. drill rod, water quenched, examined the behavior of the residual austenite in a steel of 0.70 C, 0.29% Mn. They found that about 3% austenite was retained on quenching to room temperature. If the steel were promptly chilled to -250°F, only about 0.25% austenite

remained. If, however, the quenched steel remained at room temperature for two days, 2.25% remained unchanged after chilling. The general behavior is shown in Fig. 256.

In discussion, Rose<sup>28</sup> points out that heating up to 300°F of such a steel stabilizes the residual austenite, as does standing at room

temperatures, but, if the heating is higher, up to 475°, the residual austenite disappears in 1 hr. In other steels, it is found that part of the residual austenite still remaining after supercooling will transform to bainite on heating, and further transformation of more of it to martensite occurs on the next cooling.

Hollomon, Jaffe, and Norton<sup>29</sup> discuss the effects upon transformation of remaining austenite of the presence of prior transformation of some austenite to its various decomposition products. No clear

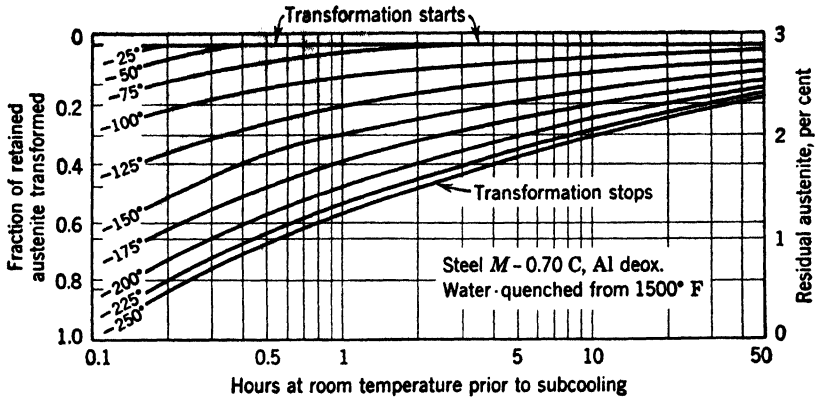


FIG. 256. Subatmospheric transformation chart for steel *M*, showing the fraction of the retained austenite that transforms and the amount of residual austenite left after room-temperature aging and subzero cooling to the indicated temperatures.

pattern emerges, but it appears that the presence of any decomposition product is likely to alter the behavior one might predict from the conventional S curve.

V. Kappel<sup>35</sup> remarks that retained austenite in aircraft parts could be transformed to martensite and the part thus become embrittled, through the low temperature reached in a high-altitude flight; hence if any doubt exists as to freedom from austenite, cold treatment prior to the final tempering might be a wise precaution.

## COLD TREATMENT

It has long been known that immersion in liquid air or, when *M<sub>f</sub>* is within the range attainable, the use of dry ice or other means of refrigeration will produce more martensite in many steels in which austenite has been retained at room temperature on an ordinary quench.

Commercial utilization of such cold treatment has sprung up especially in the treatment of high-speed tool steels, but there has been much controversy whether such treatment or repeated heating and cooling to room temperature is the better. The topic is discussed in Vol. III. The uncertainty as to  $M_f$  in many highly alloyed steels, due to the difficulty in quantitative measurement of small amounts of retained austenite, makes difficult a definite appraisal of the need for and effectiveness of cold treatment. At least, when residual austenite is suspected and there is reason to believe that it would be harmful, either in itself or by later transformation to some unwanted product, cold treatment may be logically applied.

The peculiarities appear to be tied up with what is going on in the martensite surrounding the retained austenite; tetragonal martensite changes to cubic on mild heating, then a precipitation of carbon or carbide occurs. Then the carbon becomes carbide, or the initial carbide agglomerates. Each of these changes should alter the volume and thus modify the internal pressure.

The situations attending the retention of austenite and its transformation are complicated. Unless in very highly alloyed steels, retention of large amounts occurs only in the higher-C steels. Hence, it is in the treatment of tool steels and of carburized steels rather than in ordinary forging grade steels that it calls for much consideration.

In the high-C steels, it has been noted<sup>30</sup> that the presence of undissolved carbides raises both  $M_s$  and  $M_f$ , resulting in less retained austenite.

The haziness of the picture as to the effects of retained austenite is chiefly due to inability to measure precisely the amount and location of small percentages of it. Technique for such measurements would be a boon.

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## CHAPTER 20

### SPHEROIDIZING, TEMPERING, GRAPHITIZATION

Spheroidization is the most advanced stage of tempering. It is possible to spheroidize from a variety of initial structures. By various rates of cooling from the austenitic state, different structures in different degrees of fineness are produced, as has been shown in Chapter 12. If these structures are heated, but not past  $Ac_1$ , so that no austenite is formed, their  $Fe_3C$  always tends to change toward the condition of maximum stability. The final condition of stability is decomposition of  $Fe_3C$ , into ferrite and graphite, but, except under exceptional conditions of very prolonged exposure to high temperature, the condition of apparent stability, and actual stability for ordinary conditions, is that of ferrite plus spheroidized carbide. The plates or needles of  $Fe_3C$  interspersed within the ferrite of pearlite or bainite tend to contract into balls, with less surface per volume than in the plate or needle form. It is a general chemical rule that vapor pressure and other evidences of activity are smaller as the surface-volume ratio decreases, and, as the system moves toward stability, changes tending to decrease that ratio occur. In similar fashion there is a tendency for smaller balls to coalesce into larger ones, since that action also produces a smaller surface-volume ratio.

If ferrite has been cold-worked to the threshold degree (see Chapter 9), it also tends to eliminate small crystals by their absorption into larger ones, according to the same urge to move toward a smaller surface-volume ratio. Thus, the end point of the moves toward complete stability would be one single crystal of ferrite enclosing one single globule of cementite. Diffusion of C is obviously necessary for such growth of cementite globules. The whole  $Fe_3C$  molecule does not need to diffuse, for there are plenty of Fe atoms everywhere for the C to combine with as it reaches an existing  $Fe_3C$  particle. And C does have some solubility in ferrite at the temperature being considered.

The single-ferrite-crystal one-cementite-globule condition is not obtained, because two large crystals, each eating up the small crystals

between them, reach a balance where each is as stable as the other and neither can devour the other. Unstrained ferrite has little tendency to grow larger crystals, and so the ferrite of a normalized steel retains its original crystal size while the carbide is spheroidizing. (Austenite grows large crystals without necessity for straining.)

The equally spaced  $\text{Fe}_3\text{C}$  particles, having grown to equal sizes, are also in balance; neither one is better able to attract diffusing C than the other.

Spheroidization below the critical is clearly shown in Figs. 35-37. The change in mechanical properties accompanying the change from the laminated pearlite structure to the spheroidized structure were evidenced by the data given on p. 54, Chapter 3.

**Spheroidization Temperatures.** By heating below  $A_{c3}$ , but into the austenite field, where the  $\text{Fe}_3\text{C}$  of the pearlite has not completely dissolved, or, if it has, the C-rich locations resulting from its solution have not been depleted of C by diffusion, the steel is in condition to allow easy formation of spheroidized  $\text{Fe}_3\text{C}$ , on cooling through the critical, and some spheroidization can occur above  $A_{c3}$  by the C in C-rich austenite locations diffusing up against  $\text{Fe}_3\text{C}$  particles, as the austenite cools and thus becomes supersaturated with C. Quickest spheroidization occurs when this phenomenon is utilized first, and then followed by cooling to and holding just below the critical. Since spheroidization is accomplished through the medium of diffusion of C, the shorter the distances the C has to travel, the easier the spheroidization. Hence, the finer the carbide in the original structure before heating, the better for spheroidization; that is, fine pearlite is a better starting point than coarse pearlite, bainite is still better, and so-called sorbite, resulting from tempering of martensite, is probably the best. Cold working, which breaks up  $\text{Fe}_3\text{C}$  and distributes it more uniformly, also facilitates spheroidizing. Some undissolved carbides appear necessary as nuclei.

## GRAPHITIZATION

As a final step in the approach to a completely stable structure, spheroidization can be carried to the stage where graphitization appears, the spheroidized  $\text{Fe}_3\text{C}$  breaking down to give graphite. Graphite, however, does not occur in ordinary heat treatment. Steel ordinarily acts as though it were an alloy of Fe and  $\text{Fe}_3\text{C}$ , with the  $\text{Fe}_3\text{C}$  entirely stable. In gray iron, graphite is of course present,

and in the equilibrium diagram for gray iron, fields are marked out for Fe and graphitic C as well as for Fe and  $Fe_3C$ . Malleable iron, produced by annealing white "chilled" iron to decompose  $Fe_3C$  into Fe and "temper" C, that is, finely divided graphite, shows that  $Fe_3C$  is not invariably stable.

The instability of  $Fe_3C$  in gray and chilled iron is promoted by the presence of Si, together with the high-C content, and, in recent years, "graphitic" steels have been developed, high in C and Si, in which graphite is produced from  $Fe_3C$  by suitable heat treatment.

Long ago, it was noted that high-C tool steels were occasionally subject to graphitization, but means were found to avoid this, and so little attention was thereafter given to the topic. More recently, the finding of graphite in C and CMo steels low in Si and relatively low in C, after very long use in very high-temperature steam lines, brought the disconcerting realization that, under such conditions, the  $Fe_3C$  of ordinary steels is not necessarily stable.

The controlled production of graphite in malleable iron and graphitic steels is achieved primarily by choice of composition, combined with suitable heat treatment. Its avoidance in steels where it is undesired is achieved by composition and by adhering to certain precautions in the deoxidation of the melt. These matters are discussed in Vol. II.

**Hastening Spheroidization.** Payson and coworkers<sup>1</sup> found that by starting with a relatively short sojourn at the customary spheroidizing temperature, then austenitizing at only a little above the critical, and finally dropping again to the usual temperature, spheroidization can be accelerated in the lower-C steels. They suggest the cycles given in Table 44. Short, as compared with long, austenitizing times, are emphasized.

TABLE 44

% C	1st Treatment	Austenitize	Spheroidize	Brinell
0.33	2 hr, 1340°	1 hr, 1390°	2 hr, 1350°; 4 hr, 1340°	126
0.46	2 hr, 1340°	1 hr, 1370°	8 hr, 1310°	134
0.63	2 hr, 1340°	1 hr, 1380°	8 hr, 1340°; 2 hr, 1330°	128
0.85		1 hr, 1380°	6 hr, 1335°	143
1.16		1 hr, 1400°	3 hr, 1335°	174

Brophy<sup>2</sup> emphasizes the importance of not allowing the austenitizing temperature to rise above the end of the critical range on heating, in order to secure rapid spheroidization when the temperature is

lowered, through the action of particles of untransformed ferrite. He gives the end temperatures and optimum cycles shown in Table 45.

TABLE 45

	<i>Ar</i>	<i>Ar</i> <sub>3</sub>	<i>Normalize</i>		<i>Austenitize</i>		<i>Spheroidize</i>		<i>Bri-</i> <i>nell</i>
	<i>Start</i>	<i>End</i>	<i>Hr</i>	<i>Temp.</i>	<i>Hr</i>	<i>Temp.</i>	<i>Hr</i>	<i>Temp.</i>	
T 1340	1275	1375	2	1500	2	1375	24	1225	184
2340	1225	1325	1	1475	1	1325	10	1110	202
3045	1300	1375	2	1500	2	1375	10	1250	196
3140	1275	1400	2	1450	2	1375	15	1250	174
4640	1250	1400	2	1500	2	1375	24	1200	187
4340	1275	1375	2	1550	2	1375	24	1250	195

Prenormalizing is advocated to produce grain refinement. Brophy finds that prenormalizing vastly increases tool life in machining spheroidized steels. The relatively long spheroidizing times are preferred by Brophy over Payson's shorter ones in order to give leeway to accommodate the individuality of different heats. Surprising differences in tool life are reported for different cycles producing what conventional tensile and hardness tests would indicate to be duplicate results.

However, when machinability is not the main factor, Hopkins<sup>3</sup> reports that the spheroidizing stage of annealing coils of 1038 rod for cold heading, of 3 hr at 1320–1330° after the temperature of the charge has become equalized, followed by slow cooling, reduces the properties to 67,000 tensile, 32,000 yield, 40% elong., and 66% R.A. The spheroidizing is done in controlled atmosphere in a bell-type furnace. After the holding period the heating bell is removed, and an insulated cover is put in its place for slow cooling.

The importance of precise temperature control in spheroidizing is very evident.

**Tempering of Nonmartensitic Structures.** Tempering an originally normalized structure far below the critical does not permit appreciable spheroidization and produces no visible change in the microstructure, but it does have a stress-relieving action which slightly modifies the properties. Data showing this were given on p. 411; more are given in Vol. III.

Tempering after normalizing is often applied to cast steel, producing an improvement in ductility which may mean the difference between meeting and failing to meet exacting specifications.

In tempering or spheroidizing a normalized pearlitic steel that has not been cold-worked, nothing much happens to the ferrite beyond

strain relief. Bain<sup>16</sup> remarks, "The primary purpose of tempering is to develop toughness and in no small measure this is the result of stress liberation. The stress resides in the ferrite matrix." Tempering of low-temperature bainite, and still more notably of martensite, brings about more profound alteration of structure and properties.

It should be emphasized that the mere statement that a tempered structure has a certain hardness connotes neither a particular structure nor a particular set of mechanical properties, for these depend on what the structure was before tempering, as has been brought out in reference to Jominy bars and slack quenching. As a reminder, the comparison given in Table 46, made by Jameson,<sup>4</sup> may be cited, as to the properties of 1040, when 228 Brinell has been produced by different methods.

TABLE 46

	<i>Tensile</i>	<i>Yield</i>	<i>Elong.</i>	<i>R.A.</i>	<i>Yield Ratio</i>
Normalized direct to 228	115,000	80,000	27	63	69
Fast normalize to 255, tempered to 228	122,000	91,000	25	62	74
Quenched to 600, tempered to 228	119,000	107,000	24	65	90

The greater speed with which initial softening occurs in the acicular bainite and martensite than in pearlite is shown by Engel,<sup>5</sup> Fig. 230. The behavior on longer sojourn is also shown in Fig. 229. The cause of softening is increase in particle size of  $Fe_3C$ , coupled with strain release of the ferrite. Both of these actions progress much more rapidly as the temperature is increased, but longer time at lower temperature has the same trend as shorter time at higher temperature.

**Tempering of Martensite.** Turning to the tempering of fully quenched steel, consisting essentially of pure martensite, several low-temperature phenomena occur before that of increase in particle size of  $Fe_3C$ . Indeed, it is doubtful if  $Fe_3C$  exists in freshly quenched martensite, nor is the Fe in the form of cubic ferrite, as has been mentioned on p. 50.

The quenched hardness may increase slightly in tempering up to around 300°. While  $Fe_3C$  is forming, some stress relief is taking place in the previously formed cubic ferrite. At around 400 to 500°, a medium-C steel will have lost several points in Rockwell-C hardness; the growth of  $Fe_3C$  and the regular softening has begun. Until this stage is reached, the steel is on the brittle side, especially so in

notched-bar tests. In other steels the notched-bar values may not start to rise until the tempering temperature is above 600°, as Grossmann<sup>6</sup> discusses in detail.

Quenched martensite etches light. It is the so-called "white martensite," and is difficultly distinguished from the patches of retained austenite. Light tempering makes it etch dark, perhaps from the formation of cementite from C. Tempering from about 400 to 750° brings about some coalescence of cementite and produces a structure that used to be called "secondary troostite," from its somewhat similar properties to "primary troostite," the old term applied to a nodular type of fine pearlite. Tempering at 750° up produces a still more coalesced condition of the cementite (see Figs. 33, 34).

Most steels (except the cases of carburized steels) that are heat-treated for engineering service are tempered sufficiently to be at least on the borderline between troostite and sorbite, or, in most cases, definitely sorbitic. Case-carburized steels are regularly tempered at 300–400°, with around 350° as the upper limit when file hardness is to be retained.

**Prompt Tempering.** Untempered martensite, either alpha or beta, is of little use, unless as a wear-resistant surface layer, and for this use it should be under *compressive* stress. Surface hardening by flame or induction heating under conditions such that the surface stress is compressive confers wear resistance without tempering, but in practically no other case is martensite allowed to remain untempered. The internal stress in untempered martensite is high, and, in the attempt of the steel to relieve this stress, acicular martensite is prone to crack in the quench, especially if the object quenched is irregular, or, if it survives the quench, it is prone to crack spontaneously on standing. Hence, fully quenched parts should be tempered at once without leaving the piece around at room temperature.

**Decomposition of Retained Austenite During Tempering.** Tempering in the lower-bainite region of steels with a high percentage of retained austenite, involving the production of hard bainite where the austenite used to be, may increase hardness, or at least slow down softening. Differences among steels in respect to austenite retention and to the lower parts of their S curves doubtless account for many of the differences in notched-bar impact and in R.A. at low tempering temperatures and very high hardness levels. Differences will also appear on tempering of slack-quenched steels with varying proportions of bainite, because bainite tempers more readily than martensite.

However, at ordinary hardness levels suitable for engineering use, in sections that quench fully to martensite without appreciable bainite, the properties of martensite tempered to the same hardness or tensile strength are almost indistinguishable, in spite of the presence of alloying elements in the ferrite, and in spite of the carbide not being pure  $\text{Fe}_3\text{C}$  but an isomorphous carbide containing one or several other carbide-forming elements replacing some of the Fe. (When a different carbide, not of the  $\text{Fe}_3\text{C}$  type is produced, then differences are to be expected.)

This means that it is the structure, the size, and the distribution of the carbide that rules, not the chemical composition, and that the virtue in juggling the chemical composition has been to produce hardenability in the section used, that is, to produce martensite from which the desired structure is produced by regulated tempering.

This has been shown by Figs. 68-70. As long as the specimens are longitudinal, from wrought steels, and have been fully quenched before tempering back, the plots of yield, elongation and R.A. fall right into Janitsky's original scatter bands.

**Temperability.** As has been discussed on p. 107, many alleged differences in alloy steels are claimed by using the same tempering time and temperature for each and pointing with pride to the higher strength of the one that tempers less readily or to the greater ductility of the one that tempers more readily; whereas, on the basis of tempering each to the same hardness, the properties, conventionally determined, are identical. The difference is in temperability. In some cases the steel requiring, or permitting, the higher tempering temperature is definitely to be preferred, particularly if freedom from stress is desired.

Figures 13, 146, 261 illustrate the range of strength and toughness obtainable by varying the amount of tempering to which martensite is subjected. The structures differ and, therefore, the properties differ.<sup>18</sup>

Crafts and Lamont<sup>19</sup> point out that C, B, Ni, and Mn do not confer resistance to tempering of martensite, whereas Cr, Mo, Si, and V do. Steels with a major part of their quench hardenability resulting from presence of the last four therefore require higher tempering temperatures.

**Tempering in Service.** An interesting case of highly local tempering is discussed by Jones.<sup>7</sup> Ball-bearing races, of 52100 steel, hardened to martensite and only very slightly tempered before use, finally fail in fatigue. Unless there happens to be a stress-raising inclusion



at the very surface, the nucleus of the fatigue failure is not at the surface, but a little, not quite 0.001 in. below it. It is known, from the work of Hertz and many subsequent workers, that the maximum compressive stress on the ball race, is not at, but just below, the surface.

It is contrary to expectation that repeated compressive stress should produce fatigue failure. When fatigue failure does occur under repeated compressive loads, it turns out that, as the stress is being released, a stress reversal occurs, and it is the shear stress, or the tensile component accompanying shear, that actually starts the fatigue crack. Jones so diagnoses this case. The same phenomenon occurred in the case studied by Jensen and Moore.<sup>8</sup>

Although the compressive yield strength of slightly tempered martensite is very high, service loads in very severe ball-bearing service may slightly exceed it. The energy stored in the compression cycle is not all recovered by elastic action on release of load; some plastic action occurs, accompanied by production of heat. [Like gases, metals heat when (elastically) compressed and cool when (elastically) expanded, but the heat requirements exactly balance in pure elastic action. Excess heat is generated when there is plastic deformation.]

It seems remarkable that the heat generated just below the surface should not be immediately conducted away, but, just as highly localized hot spots are produced by severe grinding of a surface (see Vol. II), so here a highly localized flash of relatively high temperature is produced at the point of maximum stress. Upon many repetitions of this flash, the originally, but slightly tempered martensite, proceeds to temper further, until finally the structure becomes visibly tempered martensite (what in older terminology was called secondary troostite). This has a lower compressive yield stress than the original structure, and it finally develops slip lines. In spite of the restraint upon permanent deformation exerted by the surrounding solid metal, this subcutaneous location manages to flow, and finally to develop a fatigue crack, analogous to those produced by over-work hardening.

The materially tempered structure and the cold-work slip lines can be revealed by metallographic methods. If the original structure contained residual austenite, as Gray<sup>12</sup> indicates is likely in 52100, this could be "isothermally" transformed to weak products; the result would be similar.

Such observations may throw a ray of light on fatigue damage. In steels not originally so hard, those tempered above the maximum

temperature attained in service, the softening would not occur but the local cold work would.

### TEMPERABILITY

The tempering time and/or temperature to produce the same hardness and strength are not the same for martensites of different steels; the alloy steels do not soften so readily. To produce 35 R.C., 160,000 psi tensile, one would temper 1040 one hour at 875°; a 0.40% C and 0.25% Mo steel one hour at 1050°. However, if in either one 35 R.C. were obtained by a longer time and a lower temperature, or a shorter time at a higher temperature, the properties would be indistinguishable; they would still fit Figs. 68-70.

**The Time-Temperature Relation.** To produce 35 R.C. from a fully quenched martensite, a particular lot of 4340 that is fairly resistant to tempering takes the time and temperature given in Table 47.

TABLE 47

10 min at 1225°
25 min at 1200
45 min at 1175
100 min at 1150
250 min at 1125
500 min at 1100
22,500 min at 1050

It would make very little difference whether this fully quenched steel were tempered 2 hr at 1140° or 20 hr at 1075°, except that more precise temperature control would be required at the higher temperature. There would be difficulty with the 10 min at 1225°, because the center of, say, a 1¾-in. section would take an appreciable part of that time before reaching 1225° and might, therefore, be tempered only to 36 R.C.; whereas, if the temperature were not well controlled and were 1250° instead of 1225°, the outside would be tempered to 34 R.C. If it were 1200° instead of 1225°, the 10-min temper would leave the outside at 36½ R.C.

At 1140° the heating lag of the center would be relatively unimportant, nor would the plus or minus 25° variation have quite so large an effect as at the higher temperature, though this is still too much. The 1225° heating in air might produce enough scaling or decarburization to be a factor. Lead-bath or salt-bath tempering would be in order if the 1225° temperature were to be used, since the

center of the piece would come to temperature more rapidly and oxidation would be avoided (compare *Metals Handbook*).

If we were dealing with a large slack-quenched section, the interior bainite of which tempers more readily than the martensitic outside, there might be virtue in choosing a high tempering temperature and so short a time that the center was at that temperature for an appreciably shorter period, if it were desired to secure approximately the same hardness throughout the section. However, the previously bainitic center, though, now having the same hardness as the previously martensitic outside, will not have exactly the same properties.

**Correlation of Time-Temperature Relations.** It is commercially desirable to cut down the tempering time as much as is compatible

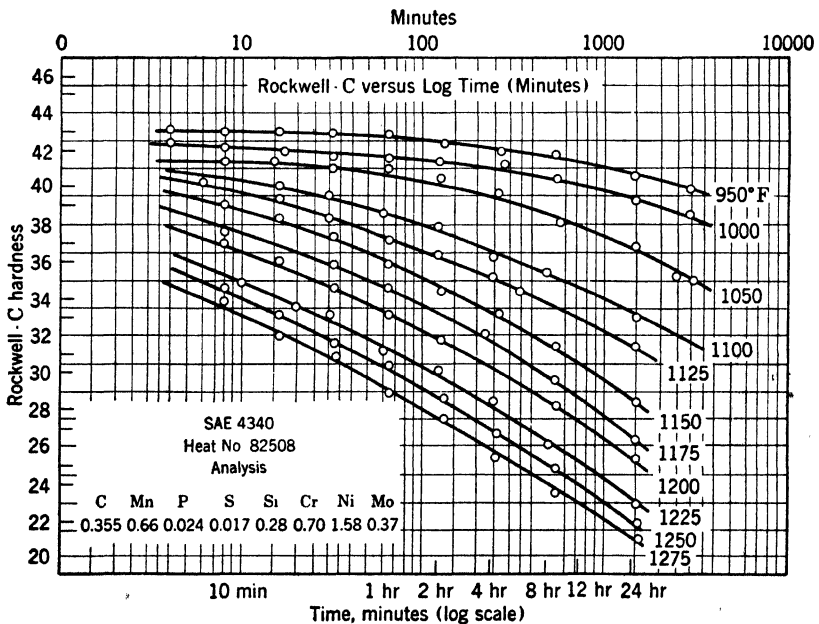


Fig. 257. Tempering at various temperatures.

with temperature equalization through the section. Because of differences in composition and other factors, in ostensibly duplicate heats, cut-and-try methods have been in vogue to determine just how to temper a given heat to secure the specified hardness in a reasonably short time.

Recent studies have indicated how the cut-and-try methods can be minimized.

Troiano<sup>9</sup> shows alternate time-temperature combinations to produce the same hardness in one lot of 4340, Fig. 257. Different heats with different temperability from this one would display a similar family of curves. Because of this similarity, a few tests of another

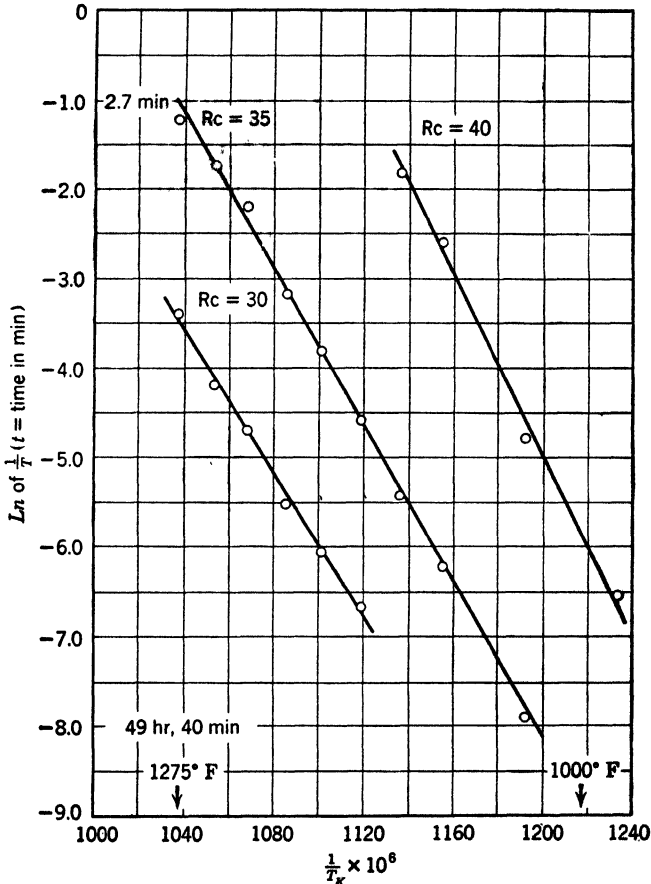


Fig. 258. Tempering relationships. Reciprocal of absolute temperature versus natural logarithm of reciprocal of tempering time.

heat, for example, tempering for 2 hr at 1050°, 1100°, and 1150° would allow plotting another curve, and its displacement from the master curve will indicate the time-temperature relations for the second heat closely enough to give a good approximation to the combinations required to temper the second heat to required hardness.

If the time-temperature combinations to produce a given hardness be plotted in terms of the reciprocal of the absolute temperature

versus the natural logarithm of the reciprocal of the tempering time, straight lines result (Fig. 258). Based on this, Troiano suggests methods of plotting by which a fairly close idea can be gained from a single tempering test of the time and temperature to give any desired hardness (Fig. 259).

A bit later, Hollomon and Jaffe<sup>10</sup> studied the time-temperature relations in tempering plain C steels, ranging from 0.31 to 1.15% C.

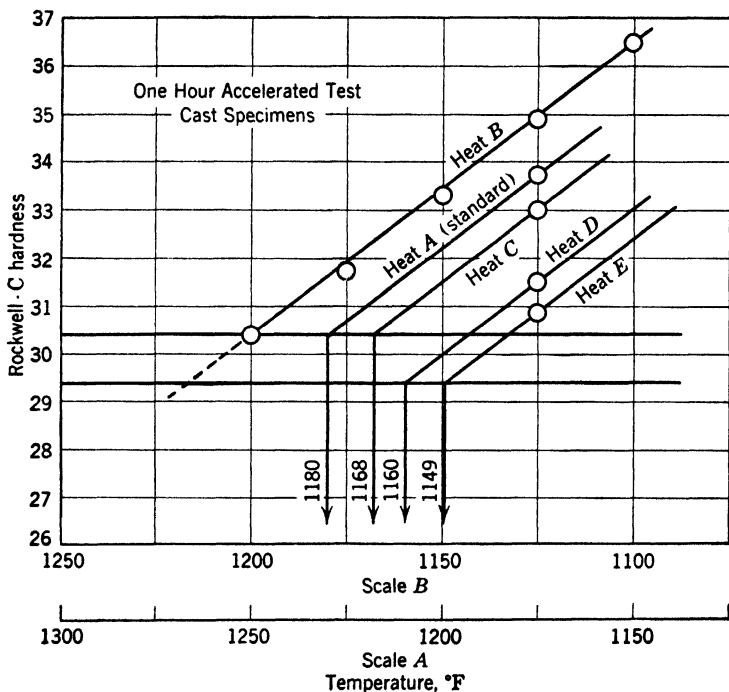


FIG. 259. Selection of tempering conditions for heats differing in temperability

Disks  $\frac{1}{4}$  in. thick,  $\frac{1}{2}$  in. in diameter, were prenormalized, austenitized, brine-quenched, and chilled to  $-95^{\circ}$  (1.15% C to  $-310^{\circ}$ ) in order to produce an all-martensite structure, which was achieved in all but the 0.31% C steel. The results are reduced to a formula analogous to Troiano's, and plots are given of hardness versus the temperature times the expression, a constant plus the logarithm of time, as in Fig. 260. The temperature plotted is in degrees Kelvin ( $^{\circ}\text{C} + 273$ ), but Fahrenheit temperatures and times have been entered in Fig. 260 for a few points. The constant used when time is expressed in seconds drops from 16 for 0.30% C to 11.5 for 1.15% C,

but using 13 as the constant introduces little error. When time is expressed in hours, the constant ranges from  $19\frac{1}{2}$  to 15. If such a master curve is known for one heat, a single tempering experiment on another heat of similar steel shows the displacement, and a curve drawn parallel to the master curve roughly predicts the tempering behavior of the second heat. The lower part of the curve is near enough to a straight line to allow drawing a parallel curve.

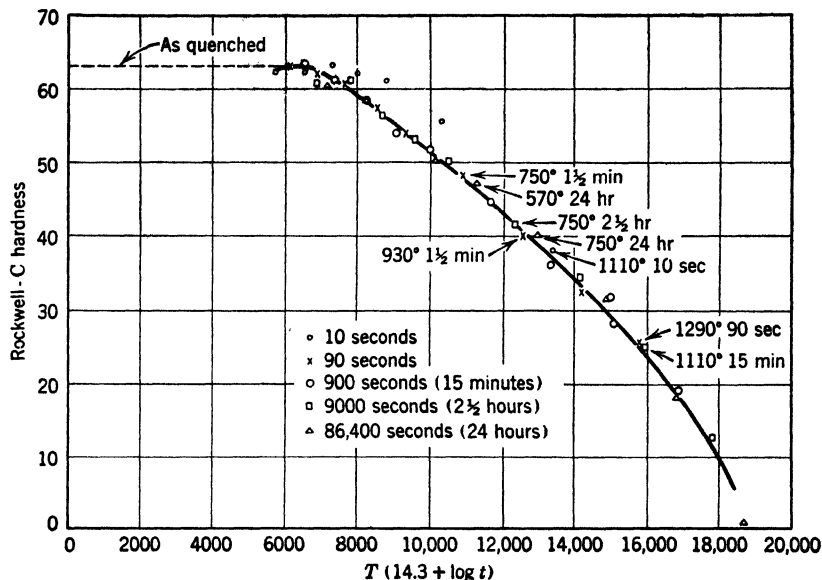


FIG. 260. Hardness versus time-temperature parameter for tempering fully quenched 0.56% C steel (S). Constant = 14.3; time  $t$  in seconds; temperature  $T$  in degrees Kelvin.

Engel's data for fully quenched 0.94% C steel, initial hardness 63 R.C., for bainite produced isothermally at  $600^\circ$  (54 R.C.) and fine pearlite produced at  $900^\circ$  (45 R.C.) are replotted by Hollomon and Jaffe on the same basis as Fig. 260. The initial softening is different for the different structures, but at hardnesses below 40 R.C. the curves do not deviate greatly. Retained austenite, secondary hardening through precipitation (see 6140 in Fig. 261 from Herres<sup>11</sup>), and graphitization during tempering of graphitizable high-C steels distort the curves. Hollomon and Jaffe give families of master curves for 0.30–0.40% and 90–1.20% C through which the behavior of a heat of steel, not affected by these features, may be roughly predicted on the basis of a single tempering experiment on that heat.

Hardenability and temperability are not necessarily related. Molybdenum adds hardenability and great resistance to tempering. Boron adds hardenability but does not affect temperability. Martensite is less dense than its decomposition products, so that a piece containing tempered martensite is smaller than it was when quenched. Selecting a composition that will temper not only to the desired

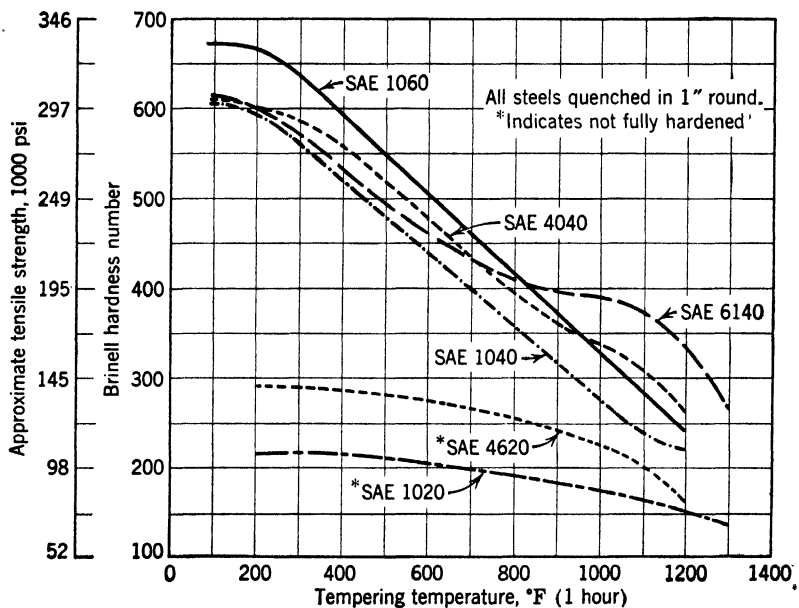


Fig. 261. Approximate tensile strength of several quenched steels as a function of tempering temperature. (Based on manufacturers' literature)

properties but also to the same dimensions it had before quenching is important in tool and die steels, as is discussed in Vol. III.

**Tempering by Stored Heat.** "Time quenching" or "quench toughening" is a term applied to some tricky processes. One of these uses a closely timed preliminary quench in water to bring the temperature close to the  $M_s$  point, followed by an oil quench to insure more gradual cooling; this leaves less internal stress in the martensite and allows production of "tough" martensite. By tough is meant a bit less brittle. If the outside has been brought below  $M_s$ , the martensite there formed may be tempered by the stored heat as it raises the temperature of the outside layer. Obviously the stored heat will vary

with the mass of the piece, so that this tempering effect is very sensitive to mass.

A process which has been used considerably for the treatment of large forgings of uniform section, such as heavy axles, is heating as usual for hardening and then quenching for a specified number of seconds, followed by air cooling. The quenching affects the steel to a certain depth, but still leaves a considerable amount of heat in the forging when removed from the bath. As the forging cools in the air, this heat from within will toughen or "soften" the steel affected by the quenching. In order to obtain equivalent results under varying conditions, the number of seconds required for immersion in the oil of a piece of given size must be determined by experiment and strictly adhered to. Forgings treated by this process are characterized by a soft or annealed core, with a progressively toughened outer part.

Rail ends may be heat-treated by spraying a regulated amount of water on the hot rail and letting the heat in the rail body temper the quenched surface layer. Whole rails have been heat-treated by timed immersion in water, and then withdrawn for tempering, in analogous fashion. Epstein<sup>13</sup> discussed somewhat analogous methods suggested by others.

A time-quenching process applied to a small symmetrical spool-shaped part of an unnamed steel, noted by Hotchkiss,<sup>14</sup> was to quench from 1450° in cold water for 10 sec, and then transfer to a salt bath at 900° which gave a surface hardness of 45 R.C., which was not considered hard enough.

However, timed quenching by a high-pressure spray of oil from all sides resulted in 54–59 R.C., and the part was used without subsequent tempering. That is, instead of fine pearlite or upper bainite, obtained by the time quenching in water, martensite was produced at the surface and then slightly tempered by the heat stored in the interior.

The varieties of time quenching have been thoroughly discussed by Burns and Brown.<sup>15</sup> The discussions of their paper, as well as its text, bring out the general undesirability of resorting to time quenching, especially because the individuality of different heats of steel is magnified under conditions of time quenching, so that even though the process is strictly controlled, the results may be erratic, and the end product is likely to contain a ring of untempered martensite, somewhere beneath the surface.



## COOLING

After tempering is completed, the piece must cool to room temperature. Even this operation is prone to introduce internal strains from purely thermal stresses.

**Slow Cooling and Stresses.** Slow cooling favors the release of internal stresses. Not only is this true of the full-annealing process, slow cooling from a temperature above that of the critical range, but also of the toughening operation. In fact, the very nature of the usefulness of toughened steel depends on the absence of a state of stress just as much as on specific static or dynamic properties. Some of the failures in locomotive forgings may be traced back to the lack of slow cooling after toughening; and this trouble sometimes justifies the requirement of cooling in the furnace, especially on very complicated objects. Just as the dangers in hardening increase with the rapidity of cooling, carbon content, and size of section, so are they likewise magnified in cooling after toughening, although on a smaller scale. If these factors become noticeably important, cooling in air from the toughening temperature may set up such a new series of cooling stresses that many of the real advantages of toughening may be invalidated.

**Use of Furnace Cooling.** The greater part of hardened and toughened work, such as automobile and other small forgings, may not require furnace cooling, apart from the fact that it may be economically impracticable. But even with such work it is desirable that the pieces be plied together after removal from the furnace so that the cooling will be retarded. For forgings of thickness greater than 3 to 4 in., such as heavy machine parts, and ordnance, cooling in the furnace is always desirable. It may be said that such cooling never did any harm, insofar as C steels are concerned, and it may do a world of good in relieving strains (compare discussion, p. 413, on normalizing large sections).

The important exception to slow cooling from the toughening temperature is that of certain CrNi, Mn, and other similar alloy steels, where temper brittleness is concerned, as has been discussed on p. 257. Lined up with the problem of internal stress is the common observation that, even though no differences in structure are discernible, long continued, relatively low temperature, tempering may lead to improved machinability or to better behavior under repeated stress than equal softening produced at higher temperature in a short

time tempering. Something analogous to creep is involved, and time is required.

**Notched-Bar Behaviors of Tempered Steel.** There is a peculiar jog in the curve of energy absorbed in notched-bar impact tests versus tempering temperature of martensite. This is shown in Armiesto's data for five heats of 4150, all falling into the composition range 0.48–0.54 C, 0.75–0.85 Mn, 0.18–0.27 Si, 0.96–1.01 Cr, 0.20–0.24% Mo, quenched in Charpy specimen size, tested untempered or tempered at 400, 500, and 600°. A nonstandard notch was ground after heat treatment.

The impact data are shown in Fig. 109. They illustrate the individuality in impact of different heats of steel, commented on in Chapter 7. Since retained austenite, later transformed to untempered martensite, might account for the 500° valley in Fig. 109, Grossmann used cold treatment and found that the valley could not be ascribed to this cause. Instead, it appears that it is related to agglomeration of carbide, at grain boundaries, which has not proceeded far enough at 400° to provide a weak uninterrupted ferrite path around the grains, but which has so proceeded at 500°. This, of course, should proceed still further at 600° and up, which would lead to the expectation of increased brittleness, rather than increased toughness, as the tempering temperature is raised further, which is contrary to fact.

The explanation is similar to that given by Grossmann (p. 182), for the brittleness of isolated ferrite patches surrounded by harder material; that is, the ferrite is so restrained that it cannot flow. Here the ferrite bands produced by 500° tempering are likewise so restrained that they cannot flow. Thus restrained, they are weak. When further tempering widens them and reduces the restraint against flow, the specimen gains toughness.

The tendency for early change at grain boundaries, before the body of the grain is affected, makes the peak and valley of Fig. 109 understandable.

**Cohesive Strength.** From the point of view of other mechanical properties, the tensile yield strength is dropping steadily as the tempering temperature increases, and so some other factor that influences toughness is acting at the hump and valley. This factor is the reaction to restraint upon plastic flow. It is given the name of "cohesive strength." Cohesive strength is not quantitatively measurable by any known direct method. More completely nonunderstandable pages have been written by highbrow authors on this topic than any

other in the realm of metallurgy and testing. Data on cohesive strength are so far from quantitative that cohesive strength has not been included in Chapter 3 as one of the special "mechanical properties."

From the behavior of metals, with various types of notches and in unnotched tests, upper and lower limits for the ratio of tensile yield strength to cohesive strength under triaxial stress can be roughly ascertained by methods such as suggested by Hollomon.<sup>17</sup> When

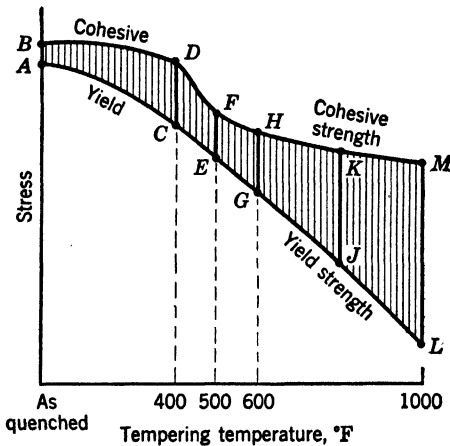


FIG. 262. Relation of yield and fracture strengths upon tempering a quench-hardened steel.

the cohesive strength falls very close to the yield strength, no matter what the level of yield strength may be, brittle behavior ensues.

A material that would be tough, that is, of high ratio of cohesive to yield strength, were it in such environment that it can deform in plastic fashion, will act brittle if it is restrained against deformation, as has been evidenced in relation to uniform materials in the discussion of the size effect in notched-bar testing. Nonuniform materials in which there is severe restraint on an otherwise tough constituent of the structure are similarly brittle.

Grossmann's Fig. 262, together with his discussion of the tempered structures and the degree of restraint exerted upon ferrite, give a good picture of the property termed cohesive strength.

**Subsidiary Tempering Effects.** Certain alloy steels display a tendency toward hardening in certain tempering temperature ranges that opposes the general tendency toward softening, thus distorting the tempering curves. This behavior is traceable to precipitation-

hardening effects, very noticeable in high-speed tool steel, and discernible in many other steels containing V or Mo. This is clearly shown in Fig. 261. Such a phenomenon has already been mentioned (Chapter 10) in the case of copper steel. The other cases are best postponed to the discussion of individual alloying elements in Vol. III.

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## CHAPTER 21

### LET'S KEEP OUR EYES OPEN

Throughout previous chapters, it has been evident that there are many misapprehensions, still insufficiently eradicated, in regard to the utility of conventional tests for evaluating steels and their heat treatment.

Nothing short of the exact type of service to which a piece is to be put is the proper foundation for evaluation. Service under static loads is one thing, that under repeated stress quite another. For the former, the importance of ductility and of "merit indices" based on ductility is frequently overestimated, and, conversely, when it is important, failure to recognize how vastly ductility may decrease, in wrought steels, between the longitudinal and the other directions, is all too prevalent.

For the case of repeated stress, the conventional polished-bar fatigue test may be misleading, since stress raisers can seldom be completely eliminated, and, when they are not, notch sensitivity becomes a vital factor. The uncertainty as to the way damage progresses under repeated stress and the lack of means for appraising its progress introduce serious uncertainties.

The conventional notched-bar impact test is prone to give very delusive data; the size effect is so marked that the usual test specimens have little value. The full-sized piece, in the exact shape it is to have in service, and under the exact service conditions, including that of temperature, should be resorted to.

As to the selection of steels, chemical composition is a much over-rated criterion, the advertising glamor, surrounding the use of one or another alloying element and the supposed virtue of adhering to some particular SAE or other steel will bear still more dispelling. There are definite uses for each alloying element, but few elements are unique in the virtues they supply. Adequate alternates can almost invariably be selected. The differences among alloy steels that seem to appear on the basis of "property charts" turn out to be pri-

marily differences in temperability, thus within the control of the heat treater.

Differences in hardenability of course exist, but, with hardenability as the criterion, a vast field of adequate alternates appears, among which, given full quenching to 100% martensite, little real differentiation can be made.

One of the most irritating misapprehensions is the one on which the 50% martensite criterion of hardenability was based, for it has led to a difficultly eradicable impression of greater depth hardening than actually exists. When the properties of a piece below the surface are of little moment, an inaccurate criterion may not matter, but, when the subsurface properties do count, the designer may be badly deceived through being led to the assumption that the center has equal properties with the surface.

In measuring hardenability by the useful Jominy test, attention to hardness only, with neglect to appraise the metallographic structure, often gives a false impression. Especially when large sections are being dealt with, the conventional relation between Jominy distance and depth hardening becomes unreliable. Further, when this relation is expressed in terms of "ideal diameter," figuring out the practical diameter, even in small sizes, is beset with uncertainties as to the intensity of the actual quench. Helpful as the Jominy test is, the best evaluation of hardenability is still to quench the actual size to be used in the actual way it is to be quenched.

Calculation of hardenability rests on so insecure a basis and so often fails to give sufficiently true values, that it is surprising that so much fruitless effort continues to be given to attempts to juggle formulae to fit some set of data, since there is no assurance that even the juggled formula will fit any other set.

As providing a mode of thinking and one means of evaluating how austenite behaves at different temperature levels, isothermal S curves are highly valuable, but they are adequate to predict the results of heat treatment only in the little-used process of austempering, which is applicable only to extremely small sections. Isothermal S curves give only the roughest sort of appraisal of behavior during continuous cooling, either in normalizing or quenching. Attempts at mathematical correction again fail to give universally reliable results. When we want to know exactly what structures and properties are produced in a given steel, in a given section, by a given method of cooling, we have to use that steel, that section, and that cooling.

Individual Jominy tests and individual S curves on one heat of steel are likely to be considered as adequately representing the behavior of another heat, supposedly identical, whereas the Jominy "curves" for a group of heats intended to be identical really form a band, and the S curves for the different heats will be considerably displaced. Again, for fully reliable information, rather than a crude approximation, a particular heat, and material from that heat, processed exactly as it is to be processed in the pieces to be heat-treated, is what should be tested.

It is not impossible to eradicate prior misapprehensions. The appraisal of grain size by the old McQuaid-Ehn test, designed for and applicable to steels to be carburized under the conditions of that test, was, in the early days, unjustifiably extended to other steels in whose heat treatment no such high austenitizing temperatures would be used. The result was serious misinformation when interpreted in terms of grain-coarsening propensities. Once the idea percolated through that, instead of the artificial conditions of a conventional test, the actual conditions of austenitization should be used in the test, true appraisal of grain-coarsening propensities ensued.

These examples tell us that short-cuts and adherence to some conventional test without thinking through the question of its applicability are fraught with danger. A test value that purports to measure a wanted quality, but actually does not, is of no value, indeed is likely to be of negative value. The more critical and more realistic attitude we take toward tests, the less likely we are to report data to the designer which mislead him, and the truer our own understanding of the principles and mechanism of heat treatment is likely to be.

A direct path to evaluation of actual behavior is more revealing, more convincing to the engineer, and very often cheaper in the long run, than the roundabout path of conventional tests.

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