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THE REAL AND IDEAL STRENGTHS OF THE COMMON STRUCTURAL METALS

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ABSTRACT

The load-bearing capacity of a structural material is, perhaps, its most important property. The reliable prediction of this property with extreme environments and under complex conditions of stress and load applications is one of the materials engineer's most formidable problems. It is all too apparent at present that materials science has not yet progressed far enough that the complex engineering properties of solids can be described in terms of fundamental quantities.

This paper briefly surveys the progress that has been made in understanding the fundamental nature of the strength of crystalline solids. The essential differences between brittle and ductile crystalline solids are first discussed. Several of the effective strengthening mechanisms of ductile solids are then reviewed. The discrepancy between the theoretical and the actual strength-weight ratios of each of the common structural metals is next examined over a wide range of temperatures. Finally, two recent high-strength iron-base alloys, designed for use in the temperature range of 0 to 1000° F, are described very briefly.

THE REAL AND IDEAL STRENGTHS OF THE COMMON STRUCTURAL METALS*

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BRITTLE AND DUCTILE CRYSTALLINE SOLIDS

Frenkel, in 1926, calculated that an ideal crystalline solid, free of macroscopic and microscopic defects, would have a fracture strength approximately equal to E/10 and a flow stress equal to G/5, where E and G are the elastic (Young's) and shear moduli, respectively.[1] His atomic model was one of close-packed spheres having attractive forces that varied sinusoidally with position. Although Frenkel's results could be viewed only as a crude estimate at best, it was apparent that the strengths of real and ideal solids differed by several orders of magnitude. In the ensuing years, much has been learned about the reasons for this startling discrepancy.

In theory, the strongest crystalline solids are those with covalent and ionic bonding. Gilman has computed the theoretical strength of several ionically and covalently bonded solids by assuming that they can be elastically strained 5%, as shown in Fig. 1.[2] Gilman concludes that it should be possible to utilize these materials at stresses in the range of several million pounds per square inch (psi), provided certain severe design restrictions are adhered to by the intended user. Among these are avoidance of surface scratches and undamped impact loads. The discrepancy between the strength of real and ideal brittle solids is now rationalized on the basis of the existence of surface or internal flaws of macroscopic or

microscopic (i.e., greater than atomic) dimensions. Thus, covalent and some ionic solids are inherently strong but inevitable surface and internal flaws initiate running cracks at applied stresses far below the theoretical strength.

Very strong, brittle materials are not likely to be used where reliability is essential or where extremely complex shapes are required. Therefore it can be concluded that although very-high-strength brittle solids would exhibit indisputable superiority in a limited number of applications, the engineering requirements of most structures will continue to demand the use of less strong but more ductile solids.

The discrepancy between the observed and theoretical strength of a ductile solid, i.e., a metal, is known to be related to the mobility of atomic-size defects called "dislocations." Dislocations are present in all metals, almost irrespective of their purity or mode of preparation. Thus, ductile metals, unlike brittle solids, are inherently weak and can be strengthened only by immobilization of dislocations that normally move under vanishingly small applied shear stresses.

STRENGTHENING MECHANISMS IN REAL METALS

A dislocation is defined as a linear defect in metals, consisting of a missing half-plane of atoms that moves under a very small applied shear stress, as shown in Fig. 2. The ease of movement of these dislocations in crystalline solids explains satisfactorily why real metals are far weaker than predicted from calculations based upon ideal or "defect-free" crystal lattices. The theory also explains why thin metal fibers or "whiskers" of many metals have strengths equal to the theoretically calculated values.

In these, the dislocations are very few, or those present are almost completely immobilized.

Dislocation theory suggests that the weakest form of a metal, viz., a high-purity single crystal, should be one in which the dislocations are relatively unimpeded by barriers, such as grain boundaries, impurity atoms, etc. Thus, high-purity single crystals and a whisker of near-theoretical strength represent the "floor and the ceiling" of the strength of a given metal, as shown in Fig. 3 for iron. The objective of the metallurgist is to increase the strength of metals from that represented by inherently weak high-purity single crystals to that exhibited by whiskers. His success in this endeavor is readily measured by comparing the strength of the alloy he has designed with that of the whisker. The highest proportions of theoretical strength reached to date in metal systems are those of iron and titanium. Some of the effective strengthening mechanisms in iron are represented in Fig. 3.

The stress required to move dislocations over appreciable distances — — i.e., to cause plastic flow — — in a high-purity single crystal is strongly dependent on the impurity content. Conservative estimates place this "flow" stress in iron single crystals of high purity at about 1,000 pounds per square inch. Conversely, the flow stress of iron whiskers has been experimentally determined to be about 1,250,000 psi. Thus, about three orders of magnitude separate the weakest and strongest forms of iron.

The most effective strengthening mechanisms of iron are those in which dislocation movement is greatly impeded or blocked. For example, when small amounts of carbon (several thousandths of one percent) are dissolved in the body-centered cubic (bcc) form of iron, the flow stress is

quadrupled. This potent strengthening is thought to be related to the positions of the carbon atoms in the iron lattice. The carbon is either "squeezed" into the holes or interstices of the lattice, or may be clustered about the dislocations, thereby creating a "drag."

Metals are crystalline solids but, unlike many naturally occurring minerals, are polycrystalline rather than monocrystalline. The transition regions (regions of crystalline "misfit") between adjacent crystals of differing orientations are called "grain boundaries." There is evidence suggesting that grain boundaries are barriers to the movement of dislocations in metals. For example, by decreasing the grain size tenfold, i.e., presenting many more barriers to moving dislocations, the strength of iron can be tripled. An electron micrograph of a "pile-up" of dislocations at a grain boundary in stainless steel is shown in Fig. 4.

Complex tangles of dislocations introduced by deforming a metal or an alloy can strongly impede the motion of other dislocations. Severe deformation of iron at room temperature doubles its strength. A transmission electron micrograph of dislocations in cold-worked metals is shown in Fig. 5.

The presence of dispersed hard particles or precipitates in alloys also serves to block dislocations. The microstructure of steel, though complex in detail, can generally be described as an aggregate of hard, brittle carbides in a relatively soft matrix of body-centered cubic iron (ferrite). The closer the spacing of these carbides and, to a somewhat lesser extent, the smaller the carbide size, the stronger will be the steel. Two well-known microstructures of steel are pearlite and bainite. The specific differences between these microstructures and others in steel is based upon the shape and distribution of the hard carbides in the ferrite, and in the manner in which the microstructures are formed. The strength of pearlite and bainite may be

increased one-third by decreasing the distance between the particles through various heat treatments. A variation of ultimate tensile strength of 120,000 to 190,000 psi for a pearlitic steel and of 186,000 to 220,000 psi for pearlitic steel of the same composition but with a bainitic microstructure is thus possible.

The strongest ductile solids known to metallurgists are the so-called "martensitic" steels. Ultimate strengths to 350,000 psi can be obtained in these quenched and tempered steels by appropriate control of composition and heat treatment. The strengthening mechanisms operative in martensitic steels are not yet completely understood in terms of dislocation theory. However, it appears that a combination of mechanisms is responsible, including several of those mentioned above, viz., solid solution and precipitation hardening.[3]

The attainable strength level for martensitic steels has recently been extended to nearly 500,000 psi by "Ausforming" — a process, developed by the Ford Motor Company, consisting of cold-working a steel prior to its transformation to martensite. Another development of both scientific and technological significance is that of the "Maraging steels" — introduced by the International Nickel Company. Their strength is not dependent upon carbon content as in conventional steels. Maraging steels have twice the ductility and many times the notch toughness of conventional steels at the same strength level. The importance of these two new materials warrants a somewhat more detailed examination of their engineering properties. This is discussed later in this paper.

Although strength levels greater than about 500,000 psi cannot be reached in bulk ductile solids, tensile strengths exceeding 650,000 psi have

been attained in fine cold-drawn steel wires (several mils in diameter). Thus, in such wires, dislocations can be so immobilized through cold-working, solution and precipitation hardening, size effects, etc., that tensile yield strengths about one-third of the theoretical whisker strength can be achieved.

The discussion has dwelt upon a comparison between the theoretical and currently attainable strength of iron at room temperature. A similar comparison, including the effect of temperature, would be of interest for all the common structural metals. Further, in such an analysis the parameter of strength-to-weight ratio rather than that of strength alone might prove to be of greater value to the designer.

REAL AND IDEAL DUCTILE CRYSTALLINE SOLIDS

A civilian advisory committee* to the Department of Defense has recently completed a comprehensive survey of the actual and the theoretical

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* The membership of Committee 2 on Metallic Materials for MAB Review of DOD Materials R & D, is as follows: Dr. Walter L. Finlay, Chairman, Crucible Steel Company of America; Dr. Spencer H. Bush, General Electric Company; Dr. Sam C. Carapella, Jr., American Smelting and Refining Company; Dr. Edward Epremian, Union Carbide Metals Company; Dr. John C. Hamaker, Jr., Vanadium Alloys Steel Company; Dr. John P. Hirth, Ohio State University; Mr. John M. Siergiej, Nuclear Metals, Inc.; Mr. George J. Wile, General Electric Company; Dr. V. F. Zackay, Ford Motor Company (on educational leave at University of California, Lawrence Radiation Laboratory, Berkeley 4, California). Special Consultants: Dr. Raymond F. Decker, International Nickel Company; Dr. Adolph J. Lena, Allegheny Ludlum Steel Company; Dr. Joseph Lane, National Academy of Sciences, Secretary.

strengths of the common structural metals.[4] The contracting agency, the National Academy of Sciences, has generously granted permission to the writer to publish the results. The writer gratefully acknowledges the efforts of Dr. Walter Finlay, Chairman, and his colleagues, who labored several years to collect and interpret the data to be shown here.

The theoretical shear stress in an ideal solid, viz., one free of defects, has been variously estimated as being between G/5 and G/15. The lower limit, i.e., G/15 is the one used herein. Following a suggestion of Dr. Hirth, a member of the above-mentioned committee, the temperature dependence of the shear modulus is assumed to be linear to about 70% of the melting point.

Since evaluation of the actual strength of the various metals is based upon tensile yield data, the value of the estimated theoretical shear strength must be doubled. Thus, the two criteria, the theoretical and actual strength-weight ratios, are $2(G/15\rho)$ and YS/ρ, respectively, where ρ is the density and YS is the measured yield strength.

The melting points, the densities, and the shear moduli of the common structural metals are shown in Fig. 6. The metals can be grouped most conveniently in terms of their melting points. These groups are Mg and Al (1202° and 1220° F, respectively); Be, Fe, Ni, and Ti (between 2332° and 3035° F); and Nb, Mo, Ta, and W (between 4474° and 6170° F).

The low densities of the elements of Mg, Al, Be, and Ti are immediately obvious, as are the high densities of the elements Ta and W. Perhaps most striking of all are the high moduli of the light metal Be and the heavy metals Mo and W.

The temperature at which the strength drops sharply is indicated in the first panel for each metal by a short horizontal line. It is apparent

that the primary factor in determining the elevated temperature strength is the melting point.

The actual and the theoretical strength-weight ratios of Al- and Mg-base alloys are shown in Fig. 7. The differences between these two ratios for Al and Mg at room temperature are by factors of about 4.5 and 8, respectively. Theoretically, then, ductile aluminum and magnesium alloys having yield strengths in excess of 300,000 psi are possible. This is in marked contrast to the highest strengths presently attainable, viz., about 100,000 psi for Al and about 40,000 psi for Mg. There do not appear to be any new developments in these alloy systems that will markedly improve their properties at room or elevated temperatures.

The principal problems of aluminum alloy design are those of improving stress-corrosion resistance, and of raising strength without concomitant loss in ductility. The most serious deficiency of magnesium-base alloys is in corrosion resistance. A better understanding of strengthening and corrosion mechanisms is needed before appreciable improvements in properties can be expected. Further, the low melting points of Mg and Al will continue to restrict their use to temperatures less than half of their melting points, i.e., to about 600° F.

The crystal structure of iron-base alloys is either body-centered cubic (bcc), face-centered cubic (fcc), or, in some instances, mixtures of both. The austenitic stainless steels are the most familiar example of those having the fcc structure, whereas the strongest alloys of the bcc structure are the martensitic ultrahigh-strength steels. The actual and the theoretical strength-weight ratios of these alloys are shown in Fig. 8.

The highest strength-weight ratios attained at room temperature for metals are those for the Ausform martensitic steels.[6] The tensile yield

strength of these alloys can be over 400,000 psi. This strength level is within a factor of about 3.5 of the theoretical yield strength. The strength-weight ratios of the austenitic stainless steels are considerably below those of the martensitic steels. The yield strengths of both austenitic and martensitic iron alloys drop sharply at temperatures above 1000° F.

The elements Ti and Be are of especial interest to the design engineer. Ti alloys, in addition to their excellent corrosion resistance, have a strength-weight ratio equivalent to those of the best ultrahigh-strength steels. Be, in addition to its light weight, has an exceptionally high shear modulus. The strength-weight ratios of Ti and Be alloys are shown in Fig. 9. Similar data for the martensitic steels are included for comparison.

Titanium alloy research and development covers a time span of less than twenty years. However, there now exist high-strength titanium alloys which are suitable for use at cryogenic, room, and elevated temperatures. Unpublished data of the Armour Research Foundation indicate, for example, that the strength of experimental Ti-Nb-Al-Zr alloys may surpass that of the Ni- and Co-base superalloys in the temperature range of 1200° to 1800° F.

The metal beryllium may well be the ultimate challenge to the metallurgist. It is relatively scarce, difficult to win from its ores, brittle, toxic in some forms, and, of course, quite expensive. On the positive side, Be has both low density and a high modulus. As a consequence of these advantages, Be has the highest theoretical strength-weight ratio of any metal. The difference between the theoretical and the actual ratios for this metal is about a factor of 40 at room temperature, as shown in Fig. 9. The great potential of Be as a structural material can be illustrated by the

following example: If Be alloys could be made with the strength of conventional ultrahigh-strength steels — — viz., about 300,000 psi — — then the strength-weight ratio of such alloys would exceed that of the theoretical ratio of either iron or titanium (50×10^5 in.).

Following the introduction of the jet turbine engine into military and commercial aircraft, the most important elevated temperature regime was between 1200° and 1800° F. The Ni- and Co-base alloys, the so-called "superalloys," were employed in this range and a great amount of research was directed toward improving their strength. The theoretical and actual strength-weight ratios for these alloys are shown in Fig. 10. For comparison, similar curves are shown for the refractory metals, Nb and Mo.

The elevated-temperature strength of the superalloys begins to fall rather steeply at temperatures above 1400° F. The prodigious research efforts in recent years aimed toward improving these alloys have resulted in only small incremental strength gains. At service temperatures above about 1700° F, the higher-melting refractory metals must be utilized.

All the refractory metals considered here have high melting points (over 4000° F), are bcc in structure, have moderate to high densities, and possess catastrophically poor oxidation resistance at their service temperatures. Reliable oxidation coatings are not yet available for most of the metals at their highest operating temperatures. This problem, and that of improving the elevated-temperature strength by alloying, without further sacrificing room-temperature ductility, are the foci of current research programs.

The theoretical and actual strength-weight ratios of Mo and Nb are shown in Fig. 10, and those of Mo, Ta, and W in Fig. 11. The elevated-temperature strengths of Mo and Nb exceed those of the superalloys above about

1800° F, as shown in Fig. 10. The "break" in elevated-temperature strength, a characteristic of virtually all metals, occurs at about 2400° F for Mo, at about 2500° F for Ta, and at about 3000° F for W. Both Ta and W have appreciable strength remaining at 3400° F.

A composite graph of the theoretical and actual strength-weight ratios, as a function of temperature for all metals so far discussed, is shown in Fig. 12. Significant improvements in the elevated-temperature properties may reasonably be expected for titanium, the refractory metals, and, perhaps, Be. The somewhat disappointing properties of the Ni- and Co-base alloys are likely to remain so until new and more effective strengthening mechanisms are discovered.

TWO NEW MATERIALS IN THE TEMPERATURE RANGE 0 to 1000° F

Two recently introduced materials designed for use in the temperature range 0 to 1000° F merit special attention. These are the high-strength iron-base Ausform and Maraging steels. The Ausform steels are of interest because they have the highest known room-temperature tensile and fatigue strengths of any ductile solid, and they combine both fabrication and heat treatment in one process.

The metallurgists, Lips and Van Zuilen, suggested about a decade ago that austenite be "cold-worked" prior to its transformation to martensite.[5] The term "cold-working" is defined here as working, or plastically deforming the austenite by a variety of fabricating techniques, such as rolling, wire-drawing, forging, etc., below its recrystallization temperature; under this treatment an entirely new set of soft strain-free grains is formed. Thus, cold-working below this temperature results in a strain-hardened or stronger austenite from which the martensite must form.

The difference between the conventional heat treatment of steel and the Dutch Process is illustrated schematically in the two panels of Fig. 13. Both hardening processes require heating to above the critical temperature in order to achieve a chemically homogeneous austenite. However, in the Dutch process, the austenite is either continuously deformed as it is cooled from the critical temperature, or, alternatively, it is deformed at a constant temperature just above the martensite "start" temperature (M_s). The deformed austenite is then quenched and tempered as in the conventional heat treatment of martensitic steels.

The early promise of the Dutch process was not realized for several years in spite of numerous unsuccessful attempts to duplicate their results. The difficulty was that the cold working of the austenite below the critical temperature caused rapid isothermal decomposition, with the result that the final microstructure was usually a mixture of coarse decomposition products. The initial encouraging results reported by Lips and Van Zuilen were on wires and thin sheets wherein the decomposition reactions were "beaten" by quick deformation and drastic cooling. It was clear that a new process was required if the advantages of this technique were to be employed in bulk steel parts having cross sections larger than wire, foil, and thin sheet. In an intensive effort, lasting about six years, researchers at the Scientific Laboratory of the Ford Motor Company succeeded in developing a process which was applicable to a wide variety of steels and fabrication techniques. This is now known as the "Ausform" process.[6]

The essential features of the Ausform process are: (a) the steel is alloyed in such a way that the undesirable isothermal decomposition reactions are suppressed over a wide temperature interval; and (b) the deformation, or cold working, is confined to this temperature interval. A comparison of this

process with the two previously mentioned is shown in Fig. 13. The steel is heated to above the critical temperature and subsequently cooled to a temperature in the relatively stable austenitic region. The ensuing deformation is performed so that the temperature of the steel does not appreciably rise or fall, thereby preventing decomposition of the austenite to either pearlite or bainite, respectively. The resulting cold-worked and strain-hardened austenite is then quenched to martensite and tempered. Many commercially available high-strength steels are of compositions that lend themselves favorably to the Ausform process. One such steel, known as Type H-11 (5.0% Cr - 1.3% Mo - 0.5% V - 0.4% C) responds especially well.

The principal factors controlling the strength of Ausform steels are the amount of deformation, the temperature at which the deformation is performed, and the carbon content. Lesser variables are the tempering temperature and the alloy content.

The optimum hardness of Ausform steels is found in a highly alloyed high-carbon steel that has been severely deformed in the austenite condition at a relatively low temperature. The mechanical properties of such a steel, called "Vascojet MA" by the producer and having 0.55% carbon and a total of 12% alloying element, are shown as a function of tempering temperature in Fig. 14. The strength and hardness (equivalent to that of a metal cutting tool) are remarkably insensitive to tempering temperature. The complex solid-state reactions responsible for the "secondary hardness peak" of the conventionally heat-treated steel are unexpectedly absent in the Ausform steel. The strength of this steel is more than 450,000 psi for a wide range of tempering temperatures, and the ductility is equivalent to its conventionally heat-treated counterpart.

The increased resistance of Ausform steels to overtempering suggests that their elevated-temperature mechanical properties might also be improved. The results of several elevated-temperature tensile tests on Ausform Vasco MA and Type H-11 steels are shown in Figs. 15 and 16, respectively. The strength of Ausform Vasco MA falls somewhat more rapidly with test temperature than does that of the conventionally treated steel, as shown in Fig. 15. However, over the range of temperatures investigated, the strength of the Ausform steel is superior to that of the conventional. For example, the strength of the Ausform steel at a temperature of 1100° F is greater than that of the conventional steel at 900° F. Further, the strength of the Ausform steel at 1000° F is equivalent to that of the conventional steel at room temperature.

The same superiority of elevated-temperature properties of the Ausform Type H-11 steel over the non-Ausform steel is noted as is observed for the Vasco MA. Of additional interest is the increase in strength and retention of ductility at cryogenic temperatures.

The endurance limits for 10^7 cycles of stress reversal were established for Ausform and conventional Type H-11, and conventional SAE 5160, as shown in Fig. 17. The endurance limits for the Ausform steel represent the highest ever recorded for any known material. There is an improvement of about 20% in endurance limit for the Ausform steel over its conventional counterpart.

Experimentally, it is known that the ratio of the fatigue endurance limit to the tensile strength of steels is about 0.5 for tensile strengths up to 250,000 psi. Above this strength level, the ratio decreases for most low-alloy steels, as shown in Figs. 18. The ratio remains high, however, for

conventionally heat treated Type H-11 steel, reflecting its excellent fatigue strength at 300,000 psi, while the Ausform steel maintains a similar ratio at a strength level of 360,000 psi.

The Ausform process combines fabrication and heat treatment in one operation. This is at once its source of strength and of weakness. The requirement that the steel be plastically deformed at a dull red heat (below about 1100° F), and that this be done over a relatively limited temperature range, currently imposes some manufacturing limitations on the process. However, utilization of more rigid fabricating equipment promises to overcome these limitations. The striking combinations of strength and ductility at cryogenic, room, and elevated temperatures, the outstanding resistance to fracture under cyclic loading, and the unusual toughness exhibited by several low-carbon Ausform steels suggest many applications for these steels -- a few of the more likely ones are shown in Table I.

The high strength, hardness, and ductility at room and elevated temperatures recommends them for tooling, such as punches, dies, shears, etc. The strength, coupled with excellent resistance to fatigue failure, has prompted studies of their use in automotive suspension systems and in high-strength bolting. Several of the highest-strength bolts ever tested were made of Ausform steel. The demonstrated high burst strength of Ausform steel tubes holds promise for application in small missile cases, mortar tubes, and rifle barrels. Preliminary ballistic tests of the low-carbon Ausform steels show them to have considerable potential as armor materials.

The discussion has been confined to steels, i.e., iron-base alloys containing carbon as the principal strengthening element. Newly introduced martensitic alloys, called "Maraging steels" by their developer -- The

International Nickel Company — do not contain significant amounts of carbon yet can be heat treated to ultimate strengths of more than 300,000 psi.[7]

A typical Maraging alloy contains 15% Ni, 9% Co, 5% Mo, 1/2% Ti, and negligible carbon. The heat treatment for this alloy is illustrated in the fourth panel of Fig. 13. The alloy is heated to above the critical temperature (about 1500° F) and allowed to air cool to room temperature. Drastic quenching is not necessary. The martensite that forms is soft and malleable, unlike that formed in quenched steels, and it can easily be fabricated into useful shapes prior to further heat treatment. The martensite is then "aged" at about 900° F for several hours. The aging process causes the soft martensite to harden. Strengths of 250,000 to 300,000 psi are readily achieved. At this strength level, the Maraging steels have about twice the ductility of conventional steel. Furthermore, their toughness is in a class by itself, as is evidenced by their ability to resist brittle fracture in the presence of sharp notches or microscopic flaws such as nonmetallic inclusions. The simple heat treatment, the excellent welding characteristics, the high toughness, and the combination of good strength and ductility at cryogenic and room temperatures have prompted a flood of development efforts directed toward early application of these steels. The suggested applications include such diverse items as highly stressed rotating parts of machinery, aircraft landing gears, hulls for deep-diving submarines, etc.

SUMMARY

Brittle crystalline solids are inherently strong but actually weak. The weakness is ascribed to macroscopic internal and surface flaws.

Ductile crystalline solids, or metals, are inherently weak but potentially strong. The weakness of metals is due to the high mobility of atom-size defects called dislocations. Metals can be strengthened by immobilizing these dislocations. Some of the most effective methods of doing this are interstitial and solid-solution strengthening, cold-working, grain refining, and precipitation hardening.

A comparison of the theoretical and the actual strength-weight ratios of the common structural metals over a wide range of temperature demonstrates the progress made in strengthening ductile solids. The discrepancy between the theoretical and the actual strength-weight ratios at room temperature varies by factors ranging from approximately 3 for Fe and Ti, to about 40 for Be. Substantial gains in strength can reasonably be expected for alloys of Ti, the refractory metals, and, possibly, Be.

The recently developed high-strength iron-base alloys, the Ausform and Maraging steels, have unusual tensile, fatigue, and notch-toughness characteristics of interest to the designer.

It is abundantly clear that from the increasing knowledge of the fundamental factors underlying the strength of solids will come a wide selection of alloys for the challenging materials problems of the future.

FOOTNOTES AND REFERENCES

* Work done under the auspices of the U. S. Atomic Energy Commission.

† On educational leave at University of California, Lawrence Radiation Laboratory, Berkeley 4, California.

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Table I. Suggested Applications of Ausform Steels

Nondefense
Advanced suspension systems
torsion bars, coil springs, single-leaf variable-cross-section springs
Tooling
punches, dies, cutting tools, shears
High-strength bolts
Aircraft parts
landing gear, structural panels, high-strength forgings
Earth-moving and agricultural equipment parts
Defense
Missile cases (especially small diameter)
Mortar and rifle barrels
Body and vehicle armor
Very-high-strength forgings, extrusions, and sheet in aerospace hardware

FIGURE LEGENDS

Fig. 1. The theoretical and currently attainable strength of some brittle solids (after Gilman).

Fig. 2. The motion of an edge dislocation and the production of a unit step of slip at the surface of the crystal. (a) An edge dislocation in a crystal structure. (b) The dislocation has moved one lattice spacing under the action of a shearing force. (c) The dislocation has reached the edge of the crystal and produced unit slip. (After Guy, Elements of Physical Metallurgy.)

Fig. 3. The influence of several compositional and microstructural variables on the strength of iron.

Fig. 4. Dislocation pile-up against a boundary.

Fig. 5. Dislocations in a cold-worked material.

Fig. 6. The melting points, densities, and shear moduli of several of the common structural metals. The short horizontal line shown in the first panel for each metal indicates the temperature at which the strength curve breaks.

Fig. 7. The theoretical and actual strength-to-weight ratios of aluminum and magnesium metals and alloys.

Fig. 8. The theoretical and the actual strength-to-weight ratios of iron and its alloys.

Fig. 9. The theoretical and the actual strength-to-weight ratios of the elements Be, Fe, and Ti and their alloys.

Fig. 10. The theoretical and the actual strength-to-weight ratios of the elements Ni, Co, Mo, and Nb and their alloys.

Fig. 11. The theoretical and the actual strength-to-weight ratios of the elements Mo, Ta, and W and their alloys.

Fig. 12. A summary of the theoretical and the actual strength-to-weight ratios of several of the common structural elements and their alloys.

Fig. 13. A schematic representation of the conventional, Dutch, Ausform, and Maraging processes.

Fig. 14. The response to tempering of the mechanical properties of Vasco MA steel in the Ausform and the conventionally heat-treated conditions.

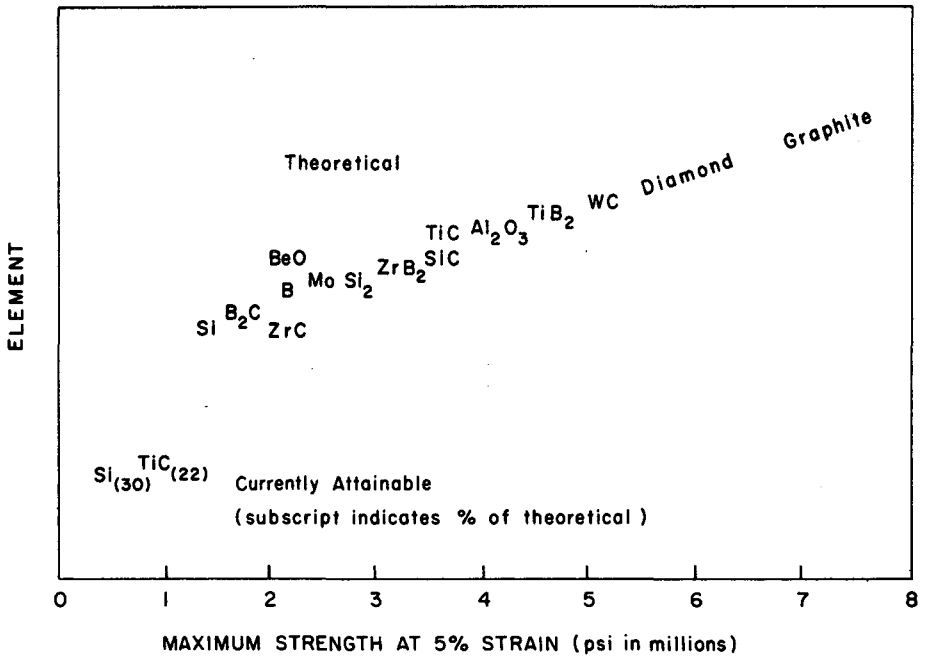
Fig. 15. The elevated-temperature strength and ductility of Ausform and conventional Vasco MA steel.

Fig. 16. The cryogenic and elevated-temperature strength and ductility of Ausform and conventional Type H-11 steel.

Fig. 17. The endurance limits of Ausform Type H-11 steel and conventional Type H-11 and SAE 5160 steels at various survival levels.

Fig. 18. The ratio of endurance limit to tensile strength for several Ausform and conventionally treated steels.

THE THEORETICAL AND CURRENTLY ATTAINABLE STRENGTH
OF SOME BRITTLE SOLIDS
(after Gilman)



MU-31062

Fig. 1.

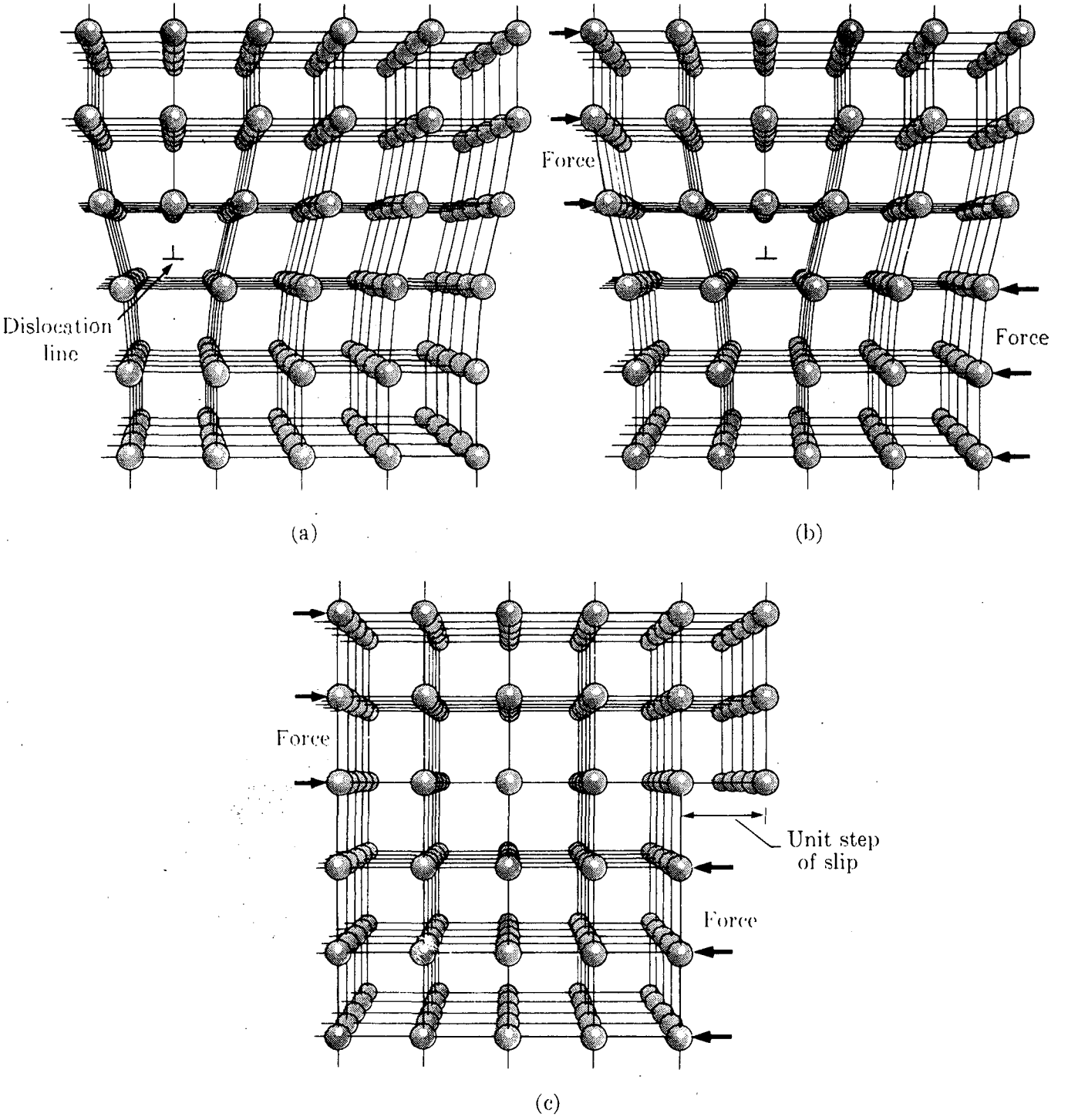
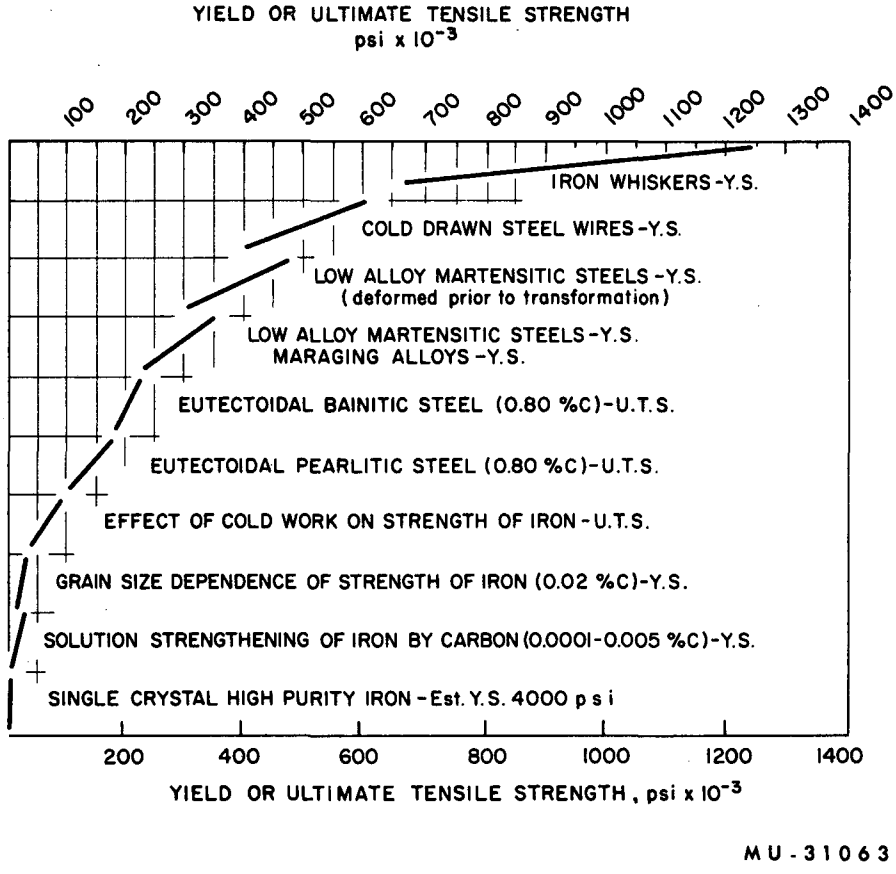
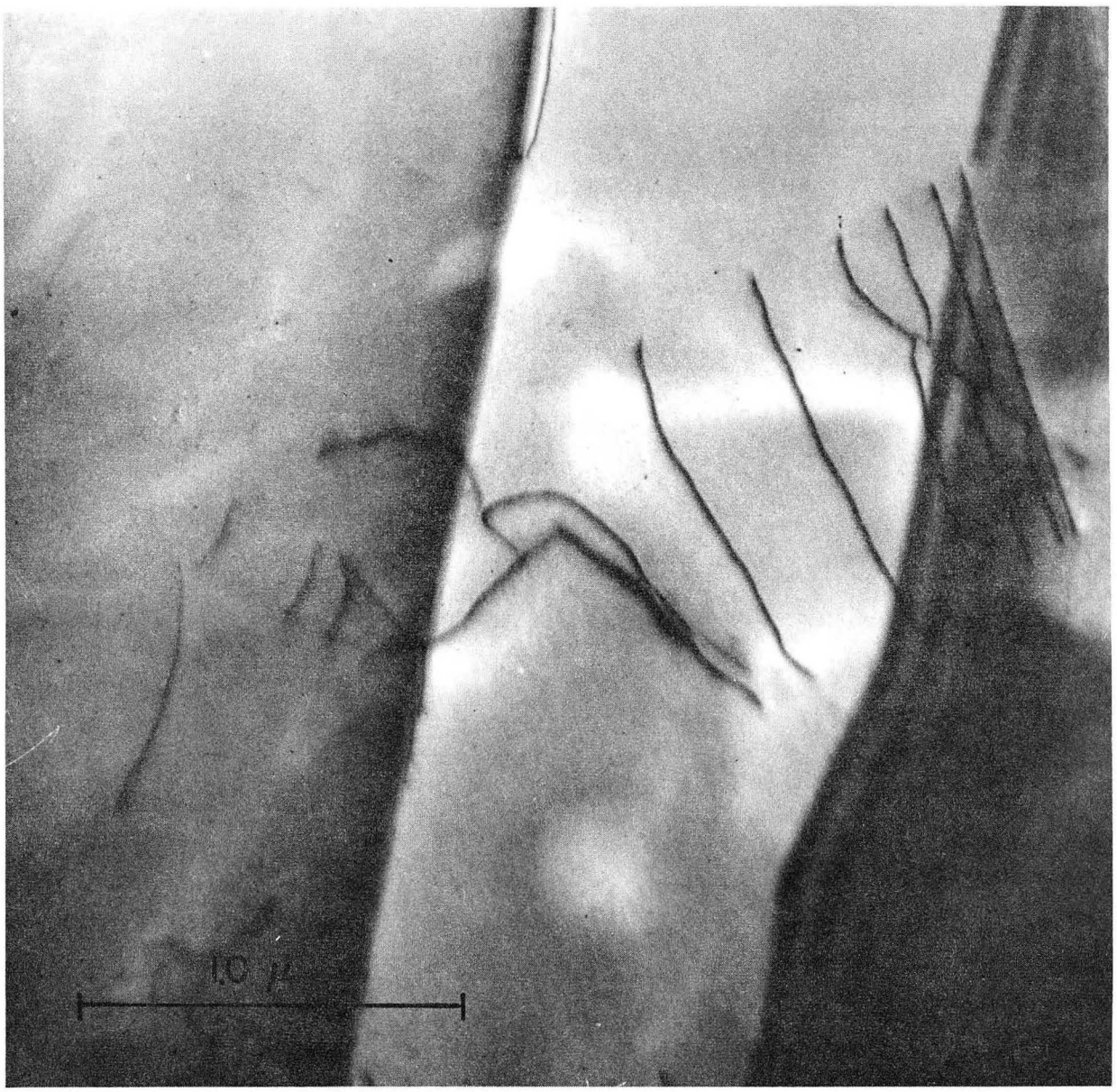

Dislocation line
Force
Force
Force
Force
Unit step of slip
(a)
(b)
(c)
MUB-1965

Fig. 2.



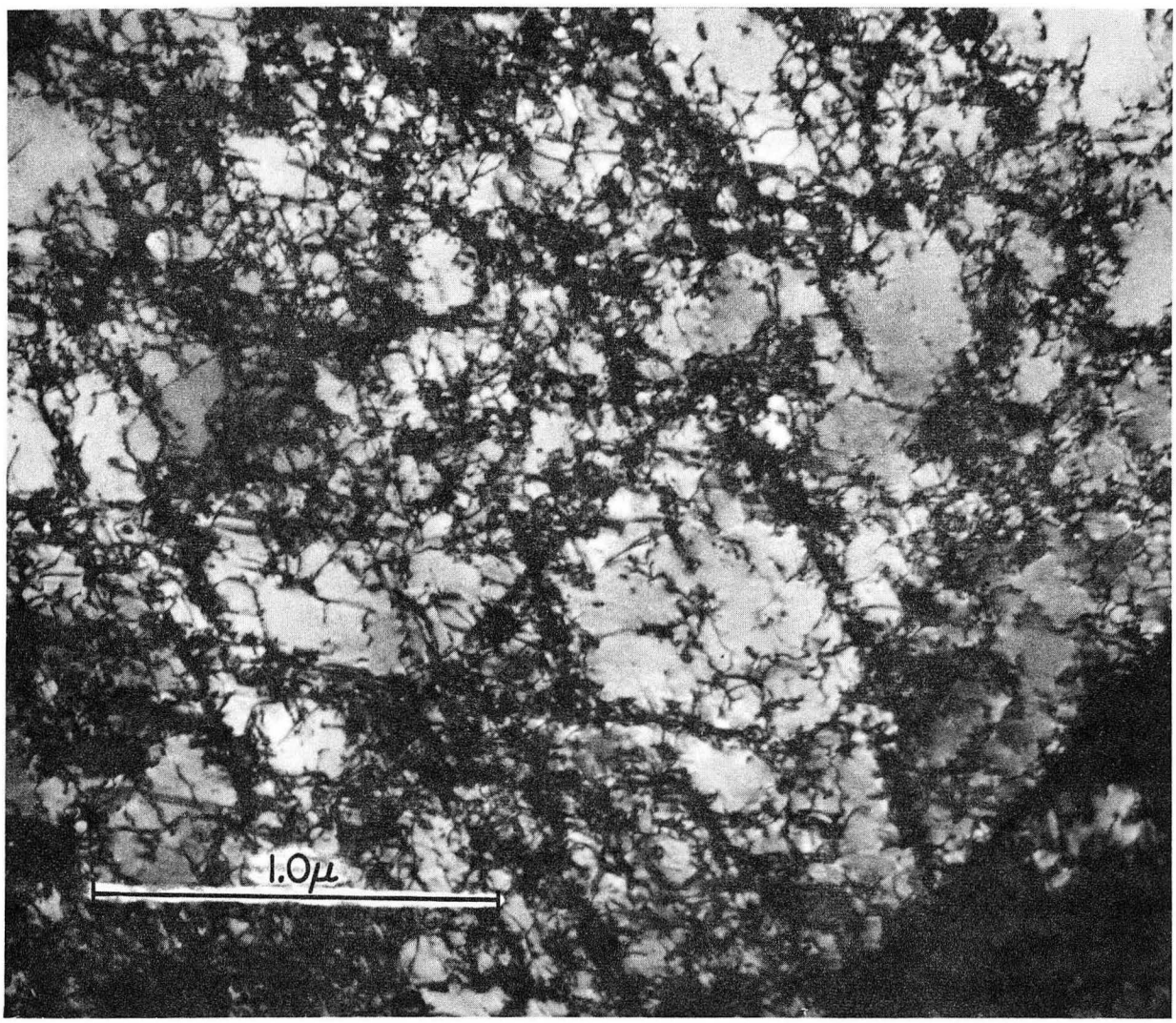
MU-31063

Fig. 3.



ZN-3836

Fig. 4.



ZN-3837

Fig. 5.

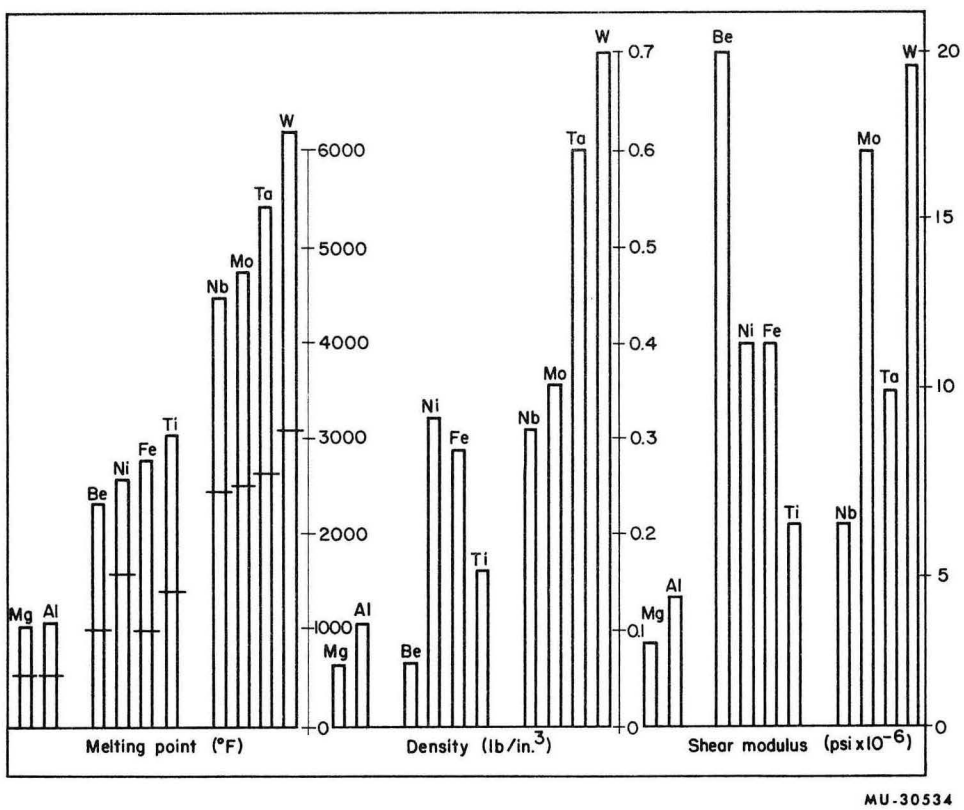

Mg
Al
Be
Ni
Fe
Ti
Nb
Mo
Ta
W
0
1000
2000
3000
4000
5000
6000
Melting point (°F)
Mg
Al
Be
Ni
Fe
Ti
Nb
Mo
Ta
W
0
0.1
0.2
0.3
0.4
0.5
0.6
0.7
Density (lb/in.3)
Mg
Al
Be
Ni
Fe
Ti
Nb
Mo
Ta
W
0
5
10
15
20
Shear modulus (psi x10-6)
MU-30534

Fig. 6.

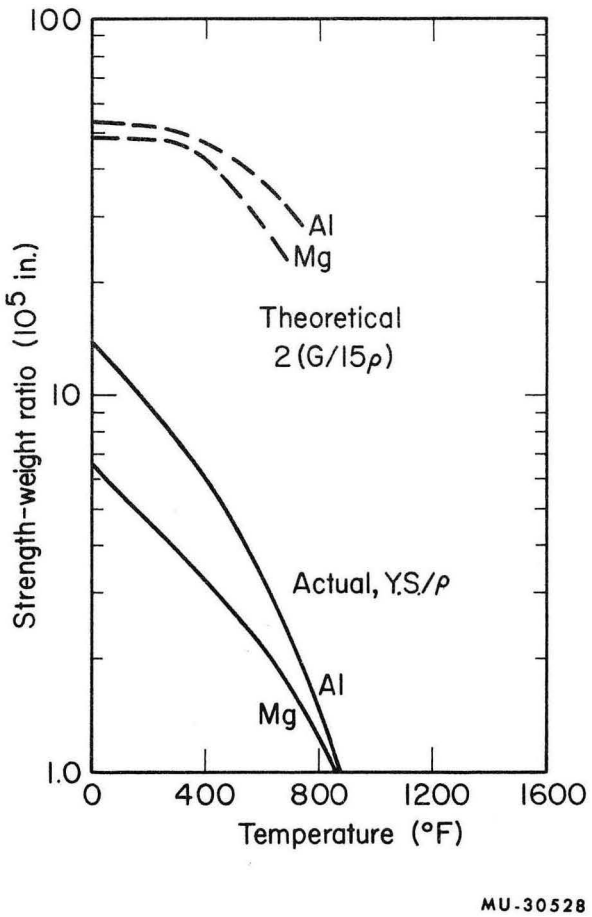

100
10
1.0
Strength-weight ratio (10^5 in.)
Al
Mg
Theoretical
2 (G/15ρ)
Actual, Y.S./ρ
Al
Mg
0
400
800
1200
1600
Temperature (°F)
MU-30528

Fig. 7.

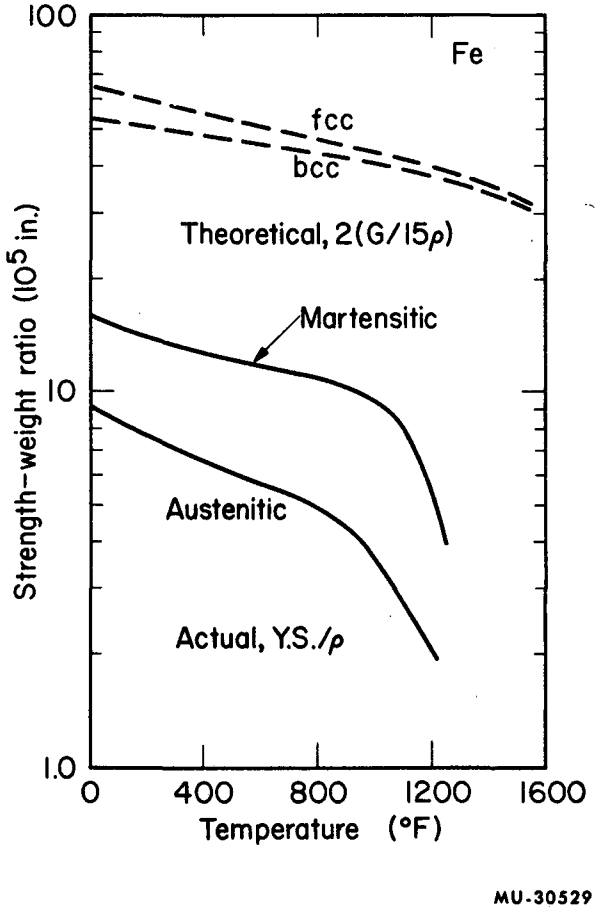

100
10
1.0
Fe
fcc
bcc
Theoretical, 2(G/15ρ)
Martensitic
Austenitic
Actual, Y.S./ρ
Strength-weight ratio (10^5 in.)
0
400
800
1200
1600
Temperature (°F)
MU-30529

Fig. 8.

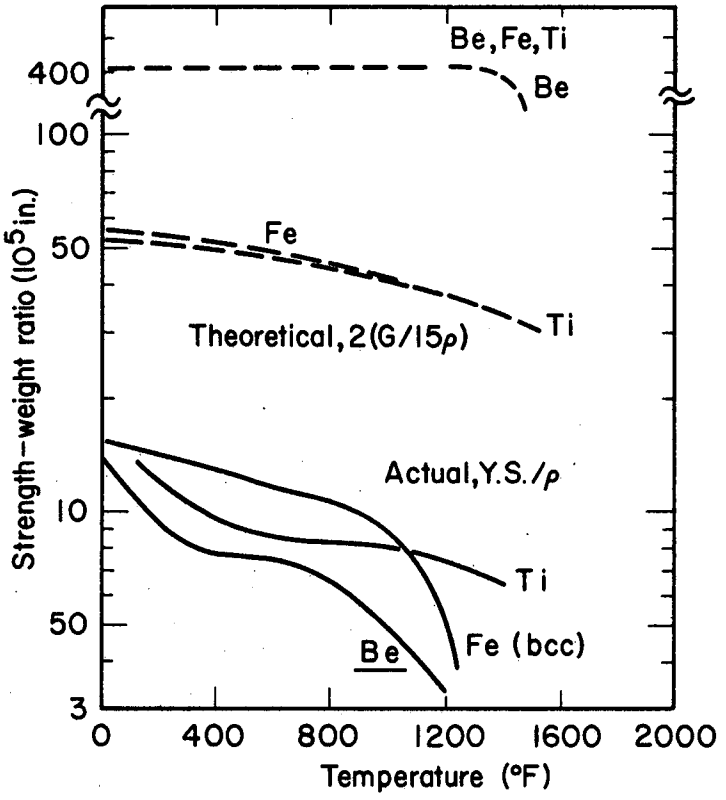

Be, Fe, Ti
Be
Fe
Ti
Theoretical, 2(G/15ρ)
Actual, Y.S./ρ
Ti
Fe (bcc)
Be
400
100
50
10
50
3
0
400
800
1200
1600
2000
Temperature (°F)
Strength-weight ratio (10^5 in.)
MU-30531

Fig. 9.

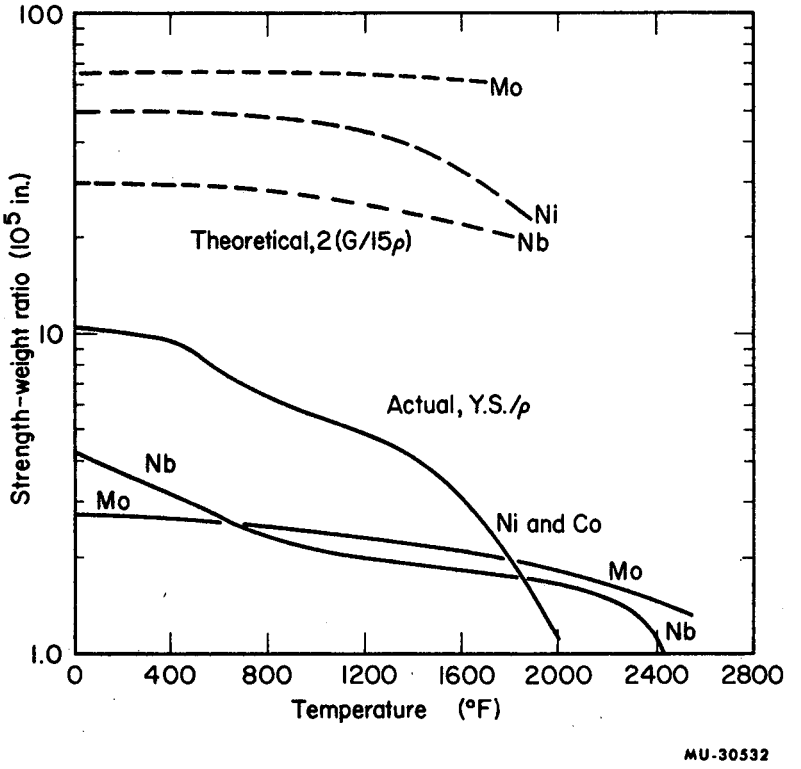

100
10
1.0
Strength-weight ratio (10^5 in.)
Mo
Ni
Nb
Theoretical, 2(G/15ρ)
Actual, Y.S./ρ
Nb
Mo
Ni and Co
Mo
Nb
0
400
800
1200
1600
2000
2400
2800
Temperature (°F)
MU-30532

Fig. 10.

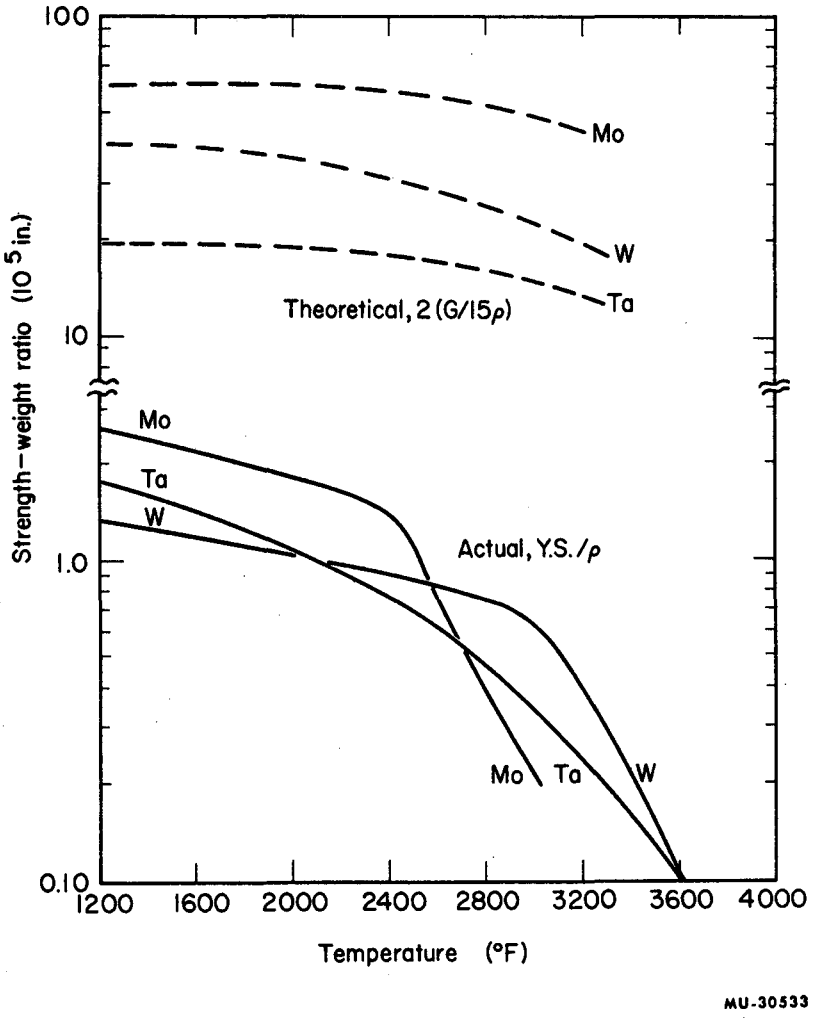
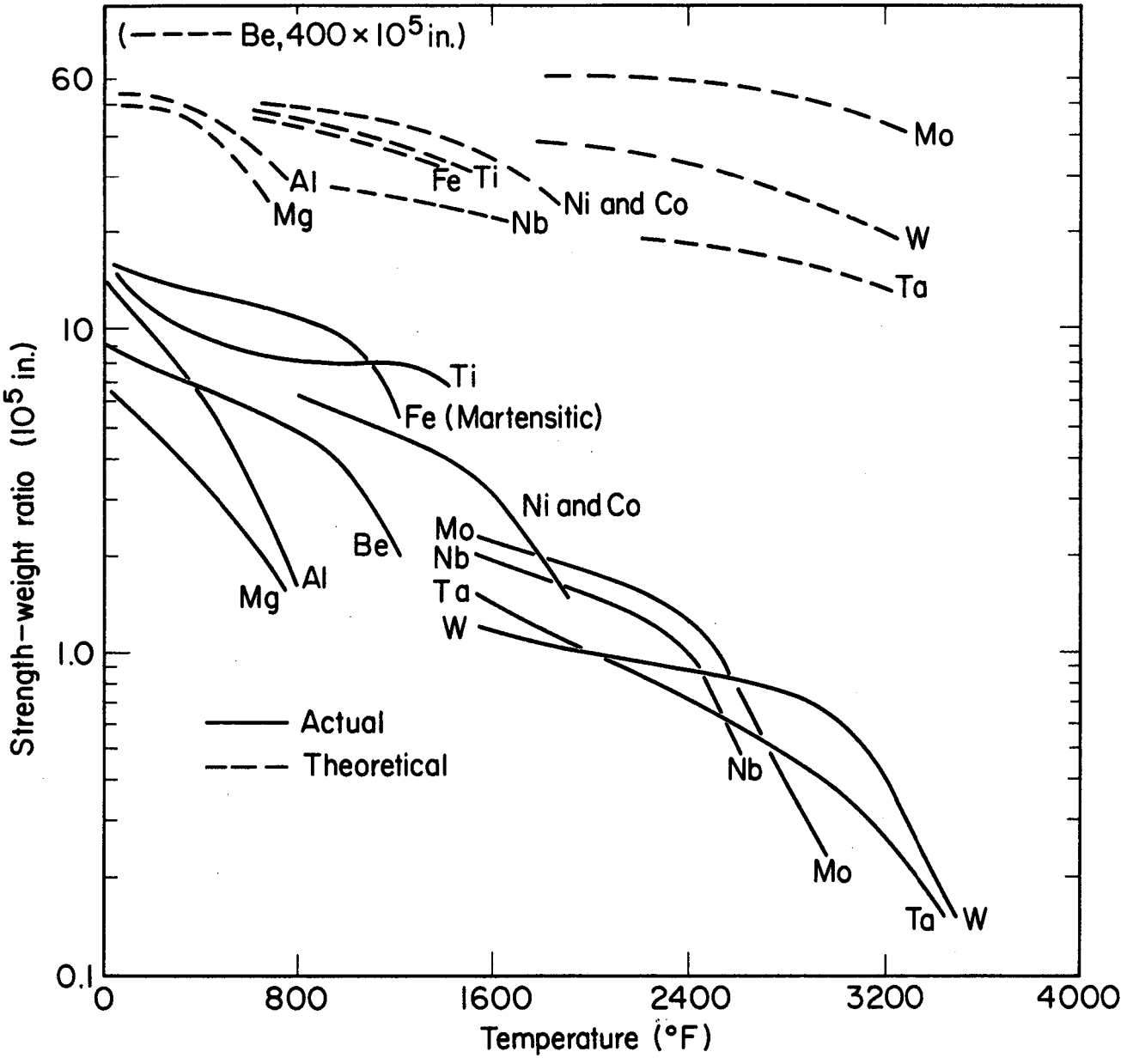

100
10
1.0
0.10
Strength-weight ratio (10^5 in.)
Mo
W
Ta
Theoretical, 2 (G/15ρ)
Mo
Ta
W
Actual, Y.S./ρ
Mo
Ta
W
1200
1600
2000
2400
2800
3200
3600
4000
Temperature (°F)
MU-30533

Fig. 11.



MUB-1833

Fig. 12.

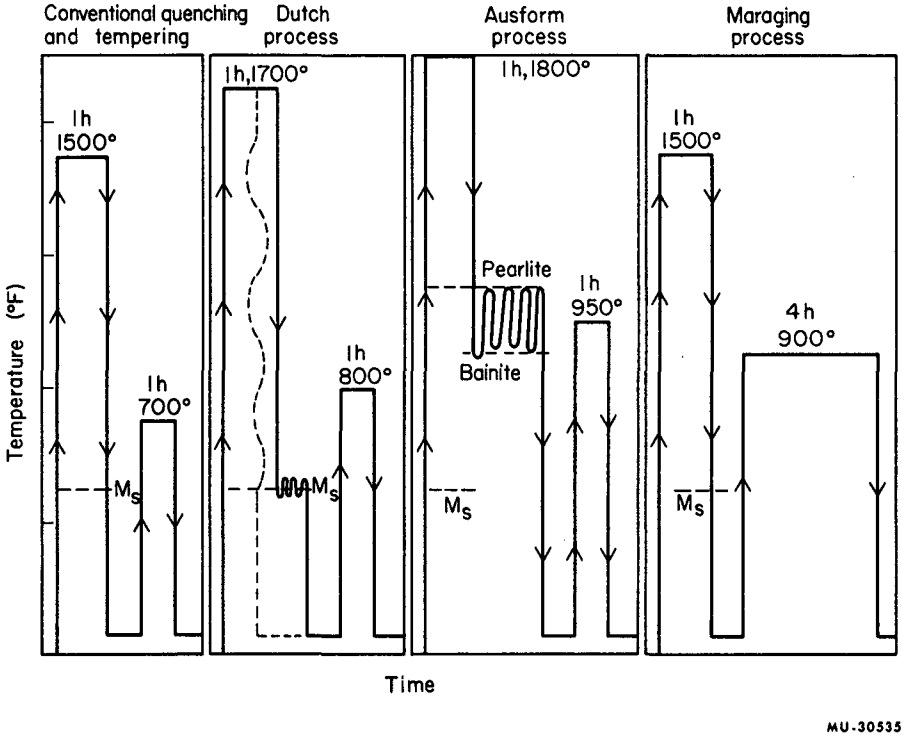

Conventional quenching and tempering
Dutch process
Ausform process
Maraging process
1h 1500°
1h 700°
M_s
1h,1700°
1h 800°
M_s
1h,1800°
Pearlite
Bainite
1h 950°
M_s
1h 1500°
4h 900°
M_s
Temperature (°F)
Time
MU-30535

Fig. 13.

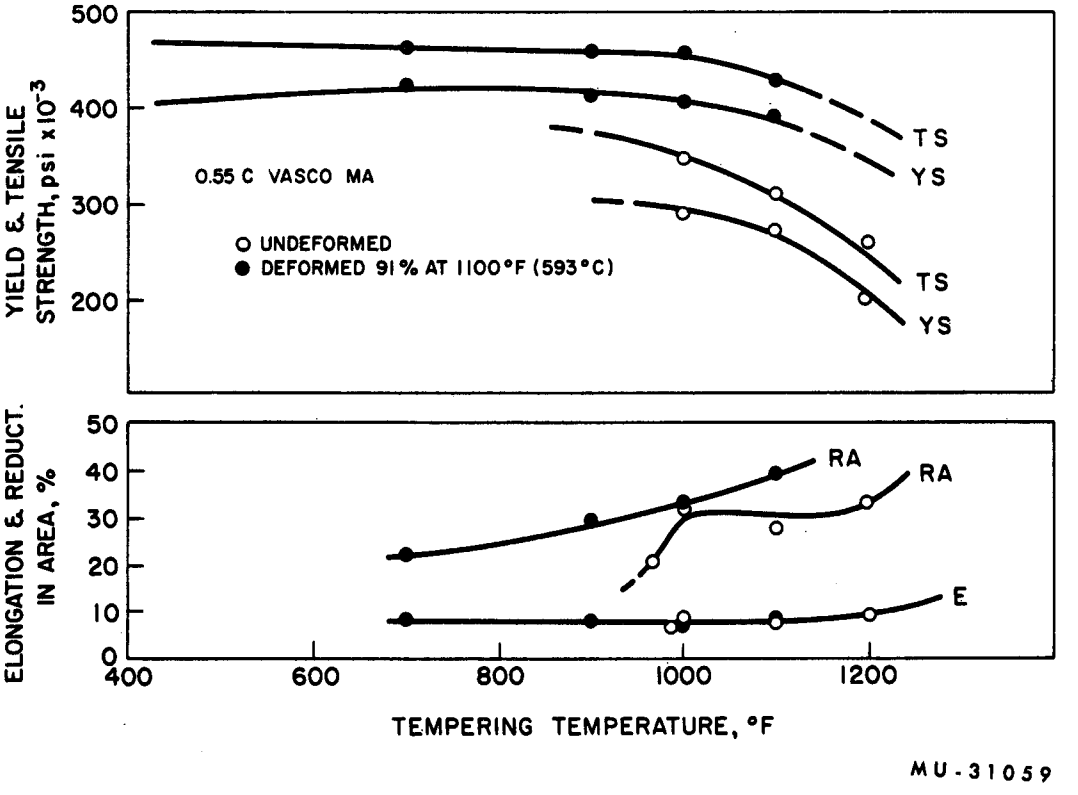
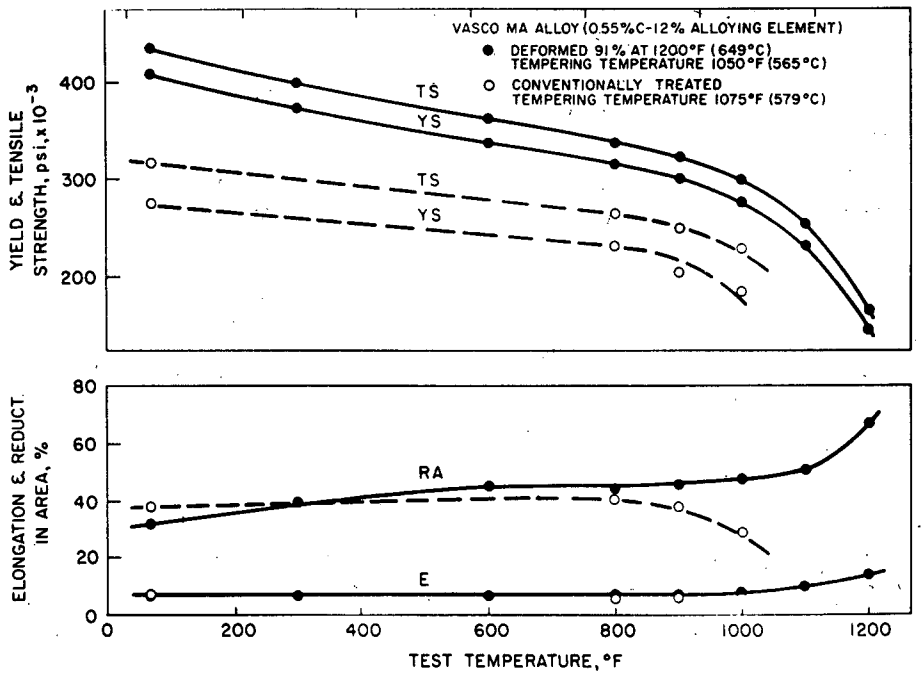

500
400
300
200
YIELD & TENSILE STRENGTH, psi x10^-3
0.55 C VASCO MA
O UNDEFORMED
● DEFORMED 91% AT 1100°F (593°C)
TS
YS
TS
YS
50
40
30
20
10
0
ELONGATION & REDUCT. IN AREA, %
RA
RA
E
400
600
800
1000
1200
TEMPERING TEMPERATURE, °F
MU-31059

Fig. 14.



MU-31057

Fig. 15.

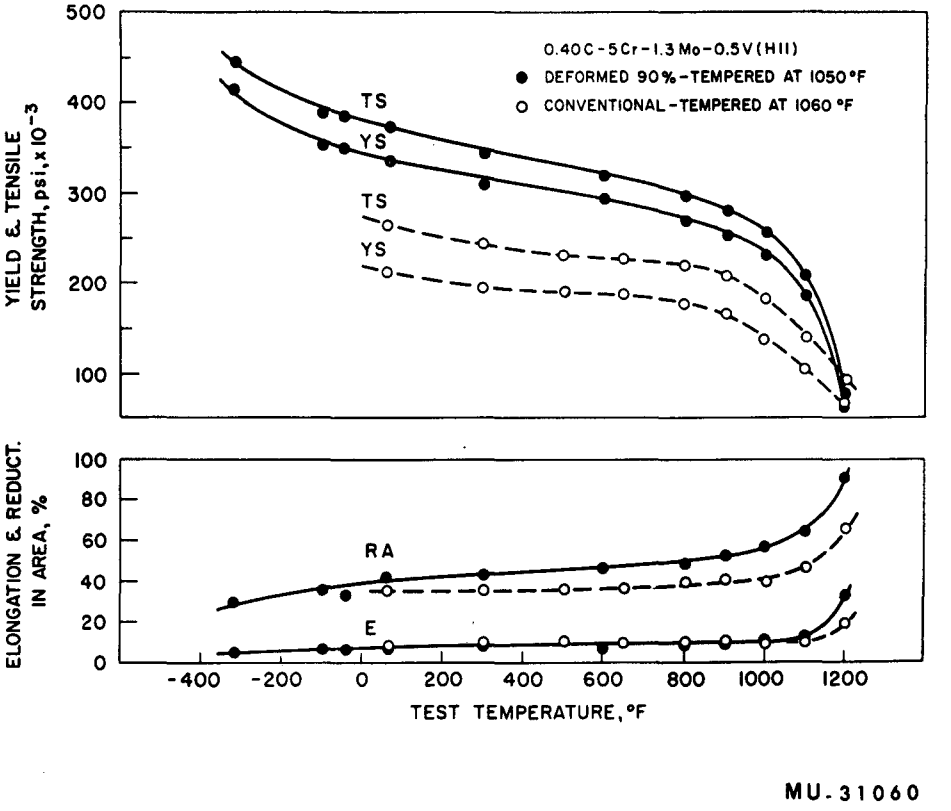

0.40C-5Cr-1.3Mo-0.5V(H11)
DEFORMED 90%-TEMPERED AT 1050°F
CONVENTIONAL-TEMPERED AT 1060°F
TS
YS
TS
YS
YIELD & TENSILE STRENGTH, psi, x 10^-3
500
400
300
200
100
RA
E
ELONGATION & REDUCT. IN AREA, %
100
80
60
40
20
0
-400
-200
0
200
400
600
800
1000
1200
TEST TEMPERATURE, °F
MU-31060

Fig. 16.

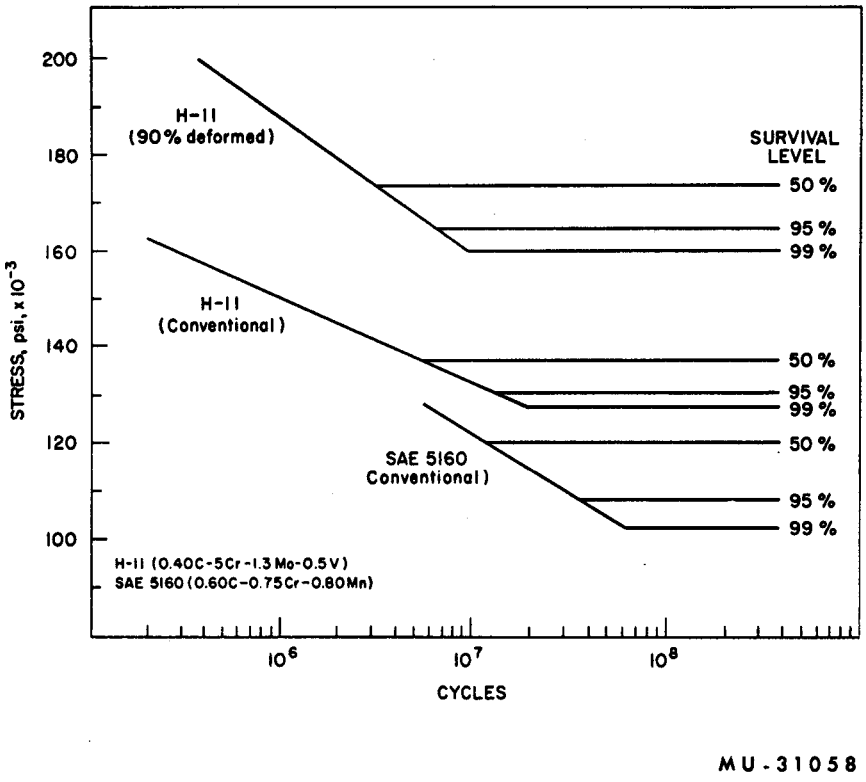

200
180
160
140
120
100
STRESS, psi, x 10^-3
H-11
(90% deformed)
H-11
(Conventional)
SAE 5160
Conventional)
SURVIVAL
LEVEL
50 %
95 %
99 %
50 %
95 %
99 %
50 %
95 %
99 %
H-11 (0.40C-5Cr-1.3Mo-0.5V)
SAE 5160 (0.60C-0.75Cr-0.80Mn)
10^6
10^7
10^8
CYCLES
MU-31058

Fig. 17.

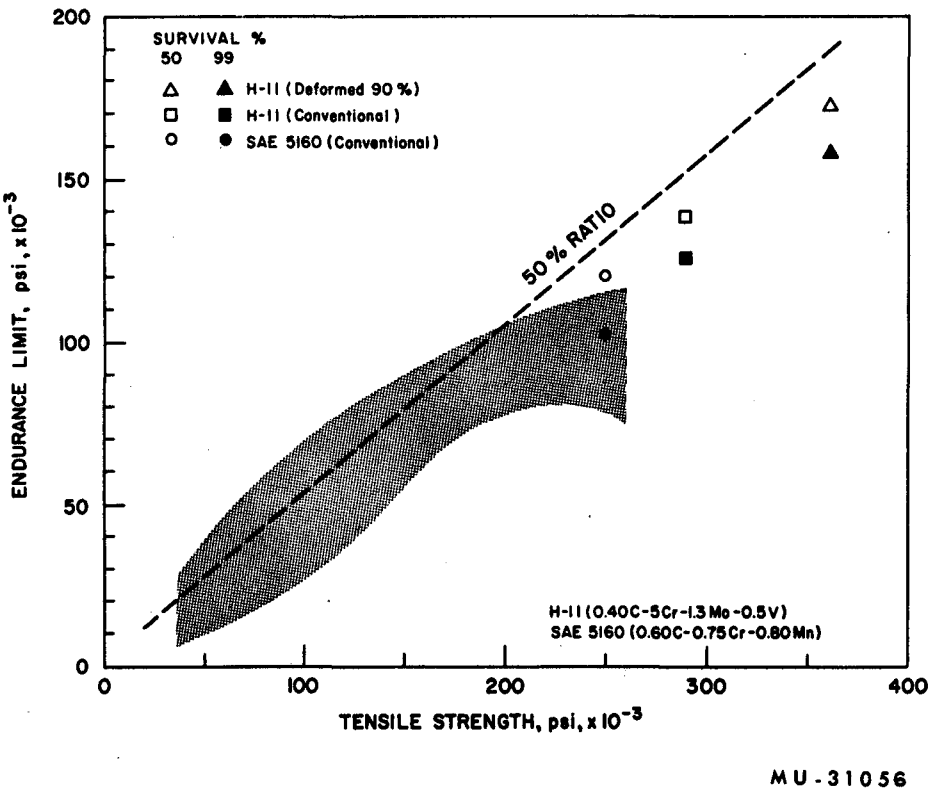

SURVIVAL %
50 99
H-11 (Deformed 90 %)
H-11 (Conventional)
SAE 5160 (Conventional)
50 % RATIO
H-11 (0.40C-5Cr-1.3Mo-0.5V)
SAE 5160 (0.60C-0.75Cr-0.80Mn)
ENDURANCE LIMIT, psi, x 10^-3
TENSILE STRENGTH, psi, x 10^-3
0
50
100
150
200
0
100
200
300
400
MU-31056

Fig. 18.

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