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WHAT WE DON'T KNOW ABOUT THERMOMECHANICAL PROCESSING

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INTRODUCTION

In the intervening ten years between the birth of the International Journal of Science and Engineering and the appearance of the commemorative edition, the field of thermomechanical treatments (TMT) has significantly changed in both scope and application. Ten years ago, much of the effort was concerned with laboratory TMT studies of ultra-high strength steels. Most of the discussion in review papers of the time dealt with the severe restrictions imposed by unacceptable processing costs and the virtually impossible fabrication problems caused by joining and equipment limitations.[1] Successful reduction of laboratory developments to commercial practice were few in number and these were usually in the military rather than in the civilian sector of technology.[2] Fortunately, the events of the past ten years have profoundly altered this unsatisfactory situation. Two major contributing factors to the renaissance of TMT have been the availability of high resolution instrumentation for the analysis of composition and microstructure, and new developments in technology; for example, the automation of shaping and forming equipment. Indeed, the most striking characteristic of the rapid technological progress made in TMT in the last ten years is the quantitative integration of physical metallurgy, viz, the structure vs property correlations, with process metallurgy, i.e., the variables of time, temperature

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and deformation.

In the ensuing discussion, the emphasis will be centered on a description of the effectiveness of this integrated analytical approach that has been used so successfully in the past several years--an approach that has culminated in the large scale production of HSLA (high strength-low alloy) steels, and one which promises to have similar technological success with respect to superalloys and, possibly, to aluminum and titanium alloys. Throughout this discussion, an attempt will be made to underscore what is not known and what additional knowledge and understanding are needed for continued progress.

DISCUSSION

In the period beginning with the Industrial Revolution and continuing well into this century, the various technological processes involving deformation and temperature were categorized according to the temperature of deformation, i.e., hot-, warm-, or cold-work. Implicit in this classification was the assumption that the structural configurations of most consequence, other than grain size and shape, were those induced well below the recrystallization temperature. In contrast, many of the new TMT processes are designed to retain defect and microstructural changes induced either above, near or below the recrystallization temperature and, in some instances, over a temperature interval encompassing many hundreds of degrees. The presence of these "tailored" residual structural configurations has a significant influence on the enhancement or retardation of the growth of minor and major phases and on the shape, size, and stability of microconstituents. By precision control of processing variables, TMT is now utilized to enhance chemical as well as mechanical properties. In addition, many TMT schedules utilize the "working" of the matrix by phase transformations rather than by shape or redundant plastic deformation. For these reasons, the older classifications of TMT by the temperature of deformation now appear to be oversimplified and somewhat outmoded, even though they are convenient and well established. In this paper, some technical highlights representative of the rapid growth of TMT in the last two decades will be described.

I. The Application of TMT to Ferrous Alloys

The recognition of TMT as a unique field of study and one of technological promise came in the mid-fifties with the publication of a short paper by the Dutch investigators, Lips and Van Zuilen,[3] wherein they described a TMT method for the production of ultra high strength steel wire. The great burst of activity following the appearance of this paper continues unabated to the present time. The progress of ferrous TMT has been summarized by a number of authors, including Kula and his associates,[1,2,4,5] May and Latham,[6,7] Latham[8], Craik,[9] Henning[10] and of the Soviet work, by Koppenaal,[11] Dunleavy and Spretnak.[12] Comprehensive biographies of the TMT literature are found in these and other reviews, in particular in the most recent one by Kula and Azrin.[2]

A. TMT Classification System

The great number and diversity of known TMT practices necessitates a system (or systems) of classification for purposes of convenience and ready comparison. It is difficult, if not impossible, to devise one scheme that accurately portrays all the TMT methods that have been proposed.

To date, two systems have come into general use--both utilize schematic representations of time-temperature-deformation schedules. Radcliffe and Kula[1] have proposed a categorization dependent upon three classes which conform to the well-known phase transformations in steel while another, following Soviet and European conventions, utilizes a similar scheme but with a different terminology. The various classes proposed by Radcliffe and Kula can be described as follows:

Class I - Deformation completed <u>before</u> the transformation of austenite, i.e., the formation of martensite from strain-hardened austenite (the ausform process is an example of a Class I TMT or LTTMT).

Class II - Deformation <u>during</u> the transformation of austenite (the formation of spherical (rather than lamellar) carbides in a ferritic matrix, called "isoforming" in Great Britain, is an example of a Class II TMT or HTTMT).

Class III - Deformation <u>after</u> austenite transformation, i.e., the strain aging of austenite transformation products (examples of TMT methods of this class are marstraining, marforming, and strain tempering or warm working).

In the European and Soviet classification scheme shown in Fig. 1, distinction is made between low temperature thermomechanical treatment (LTTMT), high temperature thermomechanical treatment (HTTMT) and sometimes a combination of the two, namely, CTMT. Araki,[13] like Radcliffe and Kula,[1] has classified the thermomechanical treatment of steel according to whether the deformation is done before, during or after a phase transformation, as shown in Table 1. It is of interest to note the variety and number of TMT's that have already been developed. Examples of the improvements in properties which can be obtained with several of these TMT's are given below.

B. Ferrous Alloys Processed by LTTMT

The significant enhancement of mechanical properties such as strength, toughness and fatigue endurance limit has been the motivating force behind the many attempts to utilize LTTMT for the production of commercial components. Although there remains some disagreement about details, most investigators agree that this enhancement of properties, especially the strength, is largely due to a combination of structural and compositional changes (relative to the conventionally treated alloy) as follows: (a) a finer martensite plate size; (b) a higher dislocation density of the martensite formed from the deformed austenite; and, finally, (c) a finer carbide size and distribution in the martensite--these carbides appear to form directly in the austenite during deformation.

Most of the TMT effort following the disclosure of Lips and Van Zuilen in the mid-fifties was with Class I or LTTMT processes. These processes are characterized by a combination of high levels (>50%) of deformation and relatively low deformation temperatures (usually between 300 and 600°C). Kula and Azrin[2] have thoroughly examined the attempts of many American investigators to reduce to practice the production of components by LTTMT methods. As previously mentioned, the only successes have been in the military sector. However, the "near-misses" in the commercial sector are interesting and the reasons for their failures are instructive. These are briefly summarized below.

One of the "near-miss" commercial cases discussed by Kula and Azrin[2] was the intended application by Bamberger[14-17] of LTTMT to the production of the inner races and ball bearings of jet engine bearings. The TTT diagram of the M-50 tool steel which was selected is shown in Fig. 2. The stippled region corresponds to the temperature interval or "bay" within which the warm working was performed. The effect of deformation on the rolling contact fatigue life is shown in Fig. 3 and the improvement of this property of the ausformed steel over that of the conventionally treated steel is shown in Fig. 4. The improvement achieved by ausforming was 650%, expressed in terms of B_{10}, the number of cycles for 10% failures. Unfortunately, although the desired properties were obtained, the increased costs occasioned by processing were not economically justified. Improvements in the melting of M-50 steel have led to significant improvements in bearing life without the need of overcoming the difficulties associated with TMT such as the critical time and temperature control required during processing, tool wear (in fabricating the parts) and high machining costs.

In the previous example of LTTMT, the strain-hardened austenite transformed after quenching to martensite (the ausform process). A variant of ausforming is the TRIP (Transformation-Induced-Plasticity) process.[18] In these alloys, the composition and processing of the alloy are adjusted such that the strain-hardened austenite does not transform to martensite on cooling. In service, the austenite transforms to martensite by a strain-induced transformation if the flow stress is exceeded. The fraction of the volume that transforms for each unit of strain depends upon the composition, the prior history (time, temperature and amount of deformation), the service temperature, and the strain rate.[19] The uniform elongation, rate of work hardening, yield strength, utlimate tensile strength and fracture toughness are all strongly influenced by the stability of the austenite with respect to strain.[19] An empirical measure of the austenite stability is the transformation coefficient "m" which is related to the strain as shown in the equation below:

$$V_\alpha = m\varepsilon^{1/2}$$

where V_α is the volume fraction of austenite transformed, and ε is the engineering strain.[20] The transformation coefficient can vary between zero (austenite completely stable to strain) to about four (highly unstable to strain). The variation of two important mechanical properties, elongation and fracture toughness (K_{IC}), with austenite stability, as measured by "m", are shown in Fig. 5 (a, b). The combinations of elongation, strength and fracture toughness that can be attained in TRIP steel are superior to those of any known alloy system.

The qualitative relationships between the processing variables of TRIP steels and several mechanical properties are shown in Table II. The numbers in the square brackets are references to the literature in the original article by Bhandarkar et al[19]. Vertical arrows indicate whether the property is increased (arrow up) or decreased (arrow down) by a corresponding change in a particular variable. In some cases, a property may change in either direction, depending on particular circumstances (for a detailed discussion of this table, see the original paper [19]).

Kula and Azrin[2] discuss several interesting applications of TRIP steels including a shear spun ring for containment of jet engine rotor fragments in the event of a catastrophic failure and of a thermally processed projectile for the radically new design of a gun barrel. In the latter, a TRIP steel was used whose strength was derived from thermally cycling through the austenite to martensite phase transformation. Koppenaal,[21,22] the inventor of this process, effectively exploited the concept of the "internal working" of an alloy by a cyclic phase transformation, as shown in Fig. 6. Unfortunately, it has been proved impossible to exceed a yield strength of about 170,000 psi. This remains an important unsolved problem. In concept, this process can be applied to weldments, castings, and low symmetry parts difficult to form by conventional processes.

Space limitations exclude, unfortunately, the description of many interesting LTTMT. These include, for example, the work of Kulin, Kreder and Kaufman[23] on the properties of steels subjected to combined TMT and cyclic heat treatments; the super-plastic behavior of high carbon pearlitic steels after warm working as shown by Sherby et al;[24] and the extensive studies by Cohen,[25,26,27] Kalish,[28] and their associates on the influence of LTTMT processes on the mechanical properties of ultra high strength steels. The recent paper by Rack and Kalish[29] is especially noteworthy as these authors show that the low stress, high cycle fatigue life of maraging steel can be improved by about 30% through the use of LTTMT. The inherent flexibility of LTTMT practices and the advantages that accrue from using them suggest that many more will be designed in the future!

C. Ferrous Alloys Processed by HTTMT

Early in the second decade of TMT, metallurgists in the steel industry began to consider the advantages of incorporating thermomechanical processing in the production of steel sheet and strip. The need for a better product was evident and the specialized production equipment had just become available. The structural steels of the time were of the hot rolled C-Mn type with typical yield strengths in the range of 40 to 50 ksi. A rising demand for strong, formable and weldable steels reflected the collective needs of the transportation and energy-related industries. In the latter category, these new steels were required for off-shore drilling rigs, powerline standards and bridges.[30] A major requirement was for vast quantities of high pressure oil pipe. The industrial steel community had largely completed a major rebuilding phase which had commenced after the close of World War II. A veritable revolution had taken place, first in Europe and then in the United States in both the refining and shaping phases of steel making. Fully automated high speed production mills that were capable of continuous precision time-temperature-deformation control and, equally important, new facilities for the rapid cooling of steel which had been deformed at elevated temperatures, were in use.

The compositional and structural requirements for a high strength low alloy (HSLA) structural steel are now well known. To achieve the strength, formability, weldability, toughness and corrosion resistance that is needed, the steel must have a fine ferrite grain size, be capable of precipitation hardening and must be free of large or elongated inclusions.[31,32,33] A typical HSLA steel contains: (i) a low carbon content to insure good weldability and toughness; (ii) one or more of the strong carbide forming elements such as V, Nb or Ti; (iii) a group of elements such as Mn, Si, . . . for solid solution strengthening and (iv) one or more of the inclusion related elements such as Zr, Ti and the rare earth metals, principally Ce and La. It is of interest to note that the reduction of grain size is the only mechanism that simultaneously increases strength, increases toughness and lowers the ductile to brittle transition temperature.[33,34]

A great variety of HSLA steels conforming to the general description given above have been successfully produced. The HSLA steels have yield strengths in the 50-80 ksi range. A major factor in the successful introduction of this new family of low alloy high strength steel has been the effective use of TMT. Some of the TMT schedules that have been employed are described in the following section.

A typical controlled rolling schedule of the type largely used in the early production of HSLA steels is shown schematically in Fig. 7.[35] After an initial rolling, the steel is held until it cools to between 1000 and 800°C and is final rolled 50-70% in the austenite range. The consequence of this TMT processing, when coupled with the use of recrystallization and grain growth retardants, such as NbC, is an extremely fine ferrite grain size. In recent years, more complex controlled rolling schedules have been employed which may involve deformation in either the γ, $\gamma + \alpha$ or α phase fields or in a combination of these. A "continuum rolling" TMT has been developed, for example, by Melloy and Dennison,[36] which provides for continuous thermomechanical treatment in all these phase fields. The essential features of several TMT processes are compared in the schematic diagram of Fig. 8. The yield strength and the transition temperature for a Charpy value of 15 ft-lbs for these processes is shown in Fig. 9. An unexpected result of the continuum rolling type TMT process was the excellent toughness at the high yield strength. These authors attribute this in part to the development of a cube-on-corner texture. The relations between the transition temperature and a parameter that describes the intensity of this texture is shown in Fig. 10. A linear relationship is shown over the entire range that was investigated.

Jenkins and Wilson[37] have shown that practical TMT schedules can be devised to produce rimmed sheet steel with a superior combination of good stretch forming properties and outstanding deep drawability. They rationalize the attainment of these properties on the basis of the development of a recrystallized texture.

It is apparent from these few brief remarks that the timely use of TMT has given the steel industry an opportunity to develop a family of low alloy construction steels with useful combinations of strength, toughness and weldability that are fulfilling vital technological needs. This success story offers some consolation to those in the metallurgical community who labored so intensely (and largely unsuccessfully) in the first decade of the existence of TMT to bring the many advantages of TMT into the commercial sector of technology.

D. What We Don't Know About Ferrous Alloy TMT

A number of investigators have attempted to summarize both the potential and the limitations of TMT. These include May and Latham,[6,7] Kula and Azrin,[2] and Craik[9] among others. All agree that the most severe barriers to technological exploitation are those associated with LTTMT. These include the excessive amount of process control that is required, the lack of reproducibility of properties and the difficulties of fabricating the final component. All of these problems plus that of joining have led to unacceptable cost penalties and to lack of commercial success. Some of these limitations are associated with the low temperatures of deformation and the large amounts of deformation that are inherent in LTTMT. These, in turn, lead to the power, lubrication and load limitations of existing shaping and forming equipment. May and Latham[6,7] cite the need for redesign of rolling and extrusion equipment. They suggest the design of new tooling and lubrication practices for existing extrusion mills and improvement of forging methods including the use of infrared heating techniques to extend the working times needed to accomplish large reductions.

The problems associated with HTTMT are more those of physical metallurgy than of equipment. HTTMT involves deformation at higher temperatures and this is comparatively easy on existing equipment. However, the relationships which exist between the microstructures and the associated mechanical properties of TMT steels are not well understood. This is especially true of the HSLA steels. The structure and associated properties of these steels are the result of a complex number of events which include the relatively unknown influence of austenite in its various recrystallized and unrecrystallized states on the nature of its decomposition products and the interrelationships of composition (at "microalloy" levels), thermomechanical treatment and transformation temperature on the final properties of the TMT steel. Relatively little is known in detail, for example of the desirable compositional and microstructural features which influence toughness and the ductile to brittle transition temperature. There is also very little information about the influence of plastic deformation on the kinetics of decomposion of austenite in low alloy steels. One of the first quantitative studies of this subject has recently been done by Walker and Honeycombe.[37] These investigators showed that the time for the start of the pearlite transformation was reduced from about 900 seconds to about 50 seconds for a 1.45% Ni, 1.1% Cr, 0.27%Mo + 0.36%C steel deformed 30% at 550°C. The effect of the amount of deformation on the transformation start for this steel is shown in Fig. 11. Increased understanding in these and other areas must be gained if the next generation of HSLA steels (yield strengths between 90 and 110 ksi) is to be successfully produced.

II. The Application of TMT to Non-Ferrous Alloys

A. Superalloys

In the past two decades, one of the most sophisticated and successful efforts of the metallurgical community has been the perfection of superalloys for gas turbine engine applications. Until recently, most of the superalloy turbine components were fabricated by conventional cast, wrought, powder metallurgical or direct solidification methods.[39] In the past several years, a number of different techniques have been developed, some of which utilize TMT. A few of these advances are briefly summarized below.

Shamblen et al[40] have devised a TMT scheme which imparts a so-called "necklace" microstructure to the superalloy René 95, as shown in Fig. 12. This microstructure exhibits superior stress rupture, notched low cycle fatigue and crack propagation resistance properties relative to those found in the more conventionally processed fine grained material. The authors ascribe the improvements to a suppression by TMT of the environmentally induced oxidation of surface carbides which in turn leads to the propagation of cracks from the surface into the interior. The retardation of this cracking is shown to be partly due to the crack barriers imposed by large warm worked grains that were formed by the TMT.

Wilcox and Clauer[41] have shown, at high fractions of the homologous temperature, that the yield stress and the creep rupture stress increase linearly with increasing grain aspect ratio (GAR), i.e., the grain length divided by the grain width. They showed that, by the use of TMT, the GAR of thoria dispersion hardened nickel alloys could be varied from 1-15 with an associated enhancement of the elevated temperature properties. In a study of the same system, Ruscoe et al[42] demonstrated that an increase of 30% in the Young's modulus could be achieved by selective TMT texturing schedules, in addition to the expected improvements in other elevated temperature properties.

Cairns et al[43] utilized a somewhat different method to attain favorable GAR's, namely, a combination of TMT and grain growth heat treatments (zone annealing). Optimization of these two processes resulted in improvements in stress rupture life by 15 to 17% over all temperatures investigated for two nickel-base superalloys.

Two of the most unusual TMT's recently developed are those of mechanical alloying[44,45,46] and thermoplastic processing. Mechanical alloying is a solid state process for producing composite metal powders with controlled microstructures. For example, nickel-base superalloy powders can be produced containing both an insoluble oxide phase Y_2O_3, and reactive alloying elements such as aluminum, titanium, and chromium. The process occurs by the repeated fracturing and welding of a mixture of metal and nonmetal powders in a highly energetic ball charge. While the overall composite particle size remains relatively coarse, tens to hundreds of micrometers, the internal structure becomes homogeneous on a submicrometer scale. Consolidation of these powders yields materials combining the stability of oxide dispersion strengthening with the corrosion resistance and intermediate temperature strength of nickel-base superalloys. Achievement of full high temperature strength in these alloys is dependent upon producing recrystallized, coarse elongated microstructures. Careful control of thermomechanical processing during consolidation, usually by hot extrusion, and secondary working by hot rolling, is required to build sufficient energy to trigger secondary grain growth to the required structures.

"Thermoplastic processing" is a P/M procedure by which superalloy powders can be consolidated and formed at 60% of the flow stress and 100°F below the temperatures used for conventional P/M proccesses. The refinement is accomplished by imparting strain energy to the powder followed by recrystallization, which occurs during the heating prior to consolidation. Atomization efficiency can be enhanced since virtually the entire output of the atomizer can be utilized. Hot isostatic compaction (HIP) of T/P processed powder can be achieved at lower temperatures and pressures; and, if disired, the HIPed compact can be superplastically formed. With thermoplastic superalloy powders, extrusion ratios on the order of 50 to 1 are possible and are expected to facilitate the production of intricate extruded shapes. T/P processed superalloy powders can be extruded at temperatures 100°F below conventional practice and hot isostatically pressed 100 to 400°F below standard HIP temperatures. In addition to the improvement in economics because of lower HIP temperatures, the as-HIPed material is a fine grained condition. This greatly facilitates subsequent conventional forging operations and also permits the application of superplastic forming techniques.

B. Aluminum Alloys

A striking feature of recent attempts to improve the properties of high strength aluminum alloys is that TMT <u>and</u> the amount and combination of alloying elements are varied in such a manner as to optimize both the mechanical properties and the corrosion resistance. Conserva and Leoni[48] demonstrated that higher yield strengths can be obtained for a given level of stress corrosion resistance by minimizing the formation of films of β or β' (Mg_2Al_3) through the TMT control of the nucleation centers and by restricting the motion of grain and subgrain boundaries. Waldman, Sulinski and Markus,[49,50] Di Russo et al,[51,52] and Thompson and Levy,[53] among others, have designed TMT's which significantly upgrade the structural properties and the corrosion resistance of several commercial alloys.

Dr. Anthony Thompson, Rockwell International, has kindly contributed, at the author's request, the following views as to the overall importance of TMT to the improvement of non-ferrous alloys:

"Among the most impressive recent advances in thermomechanical treatment or TMT is the demonstration, for both aluminum and nickel alloys, that aging prior to warm working is essential to development of both improved mechanical properties and resistance to environment fracture such as stress corrosion cracking (SCC). In aluminum alloys, an excellent example is the investigation of Paton and Sommer,[54] who were able to obtain benefits in 7075, 7049, and also in 2024. In the case of 7075, they found that aging to the peak strength or T6 condition, followed by 15% strain at 465 K and further aging 3 h at 435 K gave essentially the T6 strength with the overaged SCC properties at no sacrifice in strength or toughness from the peak condition.

This use of warm work represents a means of avoiding the planar slip characteristic of room temperature deformation. The worked structure is thus relatively homogeneous, and overaging it back to a lower strength, e.g., equivalent to the T6 condition where the process began, permits overaging the precipitates at the same time. This process is shown schematically in Fig. 13. In nickel alloys, a similar conclusion has been reached.[55,56]

Thus the refinement of substructure gained through aging prior to TMT sequences may well prove beneficial to properties of commercial aluminum and nickel alloys."

C. Titanium and Copper Base Alloys

Textures are synonomous with plastic deformation and, it is, therefore not surprising that efforts have been made to utilize TMT for the optimization of texture dependent mechanical properties.[57,58] In a recent review paper, Larson and Zarkades[57] describe the influence of texture on the mechanical properties of titanium, as shown in Fig. 14 (a-d) for the variation (with specimen orientation) of Young's modulus Fig. 14 (a); yield strength, Fig. 14 (b); impact energy, Fig. 14(c); and stress corrosion cracking (K_{ISCC}), Fig. 14(d). These authors state that virtually every important mechanical property of Ti alloys could be substantially enhanced by selecting the proper texture and orientation in the engineering component. In their view, better understanding of the relationships between composition, texture, microstructure and TMT could result in greater use of textured alloys. In some cases, microstructural features, rather than texture, account for the low properties in Ti alloys, as shown by Stubbington and Bowen[59] in their TMT study of the endurance strength of Ti-6Al-4V alloys.

Finally, one of the most intriguing TMT studies of the past few years has been made on Cu-base alloys exhibiting a spinodal decomposition. The attractive age hardening characteristics of spinodal-type alloys have been known for a long time but, unfortunately, these alloys are brittle and are, therefore, of limited structural utility. Plewes[60] succeeded in obtaining ductility at high strength levels in a Cu-9Ni-6Sn alloy with critical combinations of prior cold work, aging time and temperature. The minimum level of prior cold work required to yield the desired balance of strength and ductility was 75%. Transmission electron microscopy revealed that a continuous modulated structure develops on aging the prior cold worked material and that the periodicity of this structure was not significantly altered by the TMT. The combinations of yield strength and ductility values reported, e.g., 170,000 psi and 55% R.A., surpass those of the well known Cu-Be system. Plewes believes that these results suggest the potential, yet to be realized, in other systems exhibiting this mode of decomposition.

D. What We Don't Know About TMT and Non-Ferrous Alloys

In general, the problems associated with the application of TMT to non-ferrous alloys are unlike those of ferrous alloys, namely, they are not those involving the limitations imposed by equipment and fabrication. The principal restriction to an accelerated rate of progress is the lack of understanding of the complex interrelationships between mechanical properties and those of TMT, composition, kinetics and thermodynamics of phase transformations, microstructure and texture.

The recent studies on Ni-, Al- and Cu-base systems strongly suggest that the hitherto direct approach of superimposing plastic deformation of various amounts on the heat treat cycle of commercial alloys is no longer adequate or satisfying. There is no longer any doubt that TMT must be considered an integral part of alloy design and that many of the specialty alloys of the future will be "tailored" with respect to both composition and TMT to optimize mechanical and chemical properties.

In conclusion, it can be safely said that the progress of TMT made to date in both ferrous and non-ferrous alloys suggests that TMT is one of the most powerful tools of the alloy designer and that its full potential is yet to be realized.

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REFERENCES

1. Radcliffe, S. V. and Kula, E. B., "Deformation, Transformation, and Strength," Fundamentals of Deformation Processing, ed. by W. A. Backofen, J. J. Burke, L. F. Coffin, Jr., N. L. Reed, and V. Weiss. Syracuse: Syracuse University Press (1964), 321-63.
2. Kula, E. B. and Azrin, M., "Thermomechanical Processing of Ferrous Alloys," Advances in Deformation Processing, ed. by J. J. Burke, 21st Annual Sagamore Conference, Syracuse: Syracuse University Press (1975) in press.
3. Lips, E. M. H. and Van Zuilen, H., "Improved Hardening Technique," Metal Prog., 66, No. 2 (1954), 103-04.
4. Kula, E. B., "Strengthening of Steel by Thermomechanical Treatments" in Strengthening Mechanisms-Metals and Ceramics, ed. by J. J. Burke, N. L. Reed and V. Weiss, Syracuse: Syracuse University Press (1966), 83-121.
5. Kula, E. B. and Radcliffe, S. V., "Thermomechanical Treatment of Steel," J. Metals, 15 (1963), 755-62.
6. May, M. J. and Latham, D. J., "Thermomechanical Treatment of Steels," Part I, Metal Treating (Oct.-Nov. 1972), 3-13.
7. May, M. J. and Latham, D. J. "Thermomechanical Treatment of Steels," Part II, Metal Treating (Dec.-Jan. 1973), 3-9.
8. Latham, D. J., "The Current Position of Thermomechanical Treatments Applied to Engineering and Tool Steels," British Iron and Steel Research Association Report No. BISRA MG/A/39/69, (1969).
9. Craik, R. L., "The Effects of Warm Working on the Properties of Medium Strength Engineering Steels," Steel Strenghtening Mechanisms, Climax Molybdenum Co., Zurich, (May 1969), 45-54.

10. Henning, H. J., "Applications and Potential of Thermomechanical Treatment," Battelle Memorial Institute, Defense Metals Information Center, Columbus, Ohio, Report No. DMIC-MEMO-251, (Nov. 1970).

11. Koppenaal, T. J., "The Current Status of Thermomechanical Treatment of Steel in the Soviet Union," ASM Trans.Quart., 62 (1969), 24-37.

12. Dunleavy, J. G. and Spretnak, J. W., "Soviet Technology of Thermal-Mechanical Treatment of Metals," Battelle Memorial Institute, Defense Metals Information Center, Columbus, Ohio, Report No. DMIC-MEMO-244, (Nov. 1969).

13. Araki, T., Toward Improved Ductility and Toughness, Climax Molybdenum Co., (1971), 358-359.

14. Bamberger, E. N., "The Effect of Ausforming on Rolling Contact Fatigue," Final Engineering Report, Bureau of Naval Weapons Contract No. NOw-63-0144-d, (December 1963).

15. Bamberger, E. N., "The Effect of Ausforming on the Rolling Contact Fatigue Life of a Typical Bearing Steel," Trans. ASME, Ser. F, J. Lub. Tech., 89 (1967), 63-75.

16. Bamberger, E. N., "The Production, Testing, and Evaluation of Ausformed Ball Bearings," General Electric Co., Flight Propulsion Division, Cincinnati, Ohio, Final Engineering Report, Bureau of Naval Weapons Contract No. NOw-65-0070-f,(June 1966) AD 637576.

17. Bamberger, E. N., "Manufacture of Jet Engine Thrust Bearings by Ausforming," General Electric Co., Aircraft Engine Group, Cincinnati, Ohio. Air Force Materials Laboratory Contract Report No. AFML-TR-72-43, (April 1972) AD 894156L.

18. Zackay, V. F., Parker, E. R., Fahr, D., and Busch, R., "The Enhancement of Ductility in High Strength Steels," ASM TRANS QUART., 60 (1967), 252-59.

19. Bhandarkar, D., Zackay, V. F., and Parker, E. R., "Stability and Mechanical Properties Of Some Metastable Steels," Met. Trans. 3, (1972), 2619.

20. Gerberich, W. W., Thomas, G., Parker, E. R., and Zackay, V. F., "Metastable Austenites: Decomposition and Strength," in Proc. Second Int. Conf. on The Strength of Metals and Alloys, Asilomar, California, August 1970, 894-99.

21. Koppenaal, T. J., "A Thermal Processing Technique for TRIP Steels," Met. Trans., 3 (1972), 1549-1554.

22. Koppenaal, T. J., "Thermal Processing of TRIP Steel," paper presented at Colloquium on TRIP steels, Army Materials and Mechanics Research Center, Watertown, Mass., (June 1974), 27-28.

23. Kulin, S. A., Kreder, K., and Kaufman, L., "Effect of Cyclic Transformation Treatments on the Structure and Mechanical Properties of Steels," Final Report, Office of Naval Research, Contract No. N0014-70-C-0039, NR 031-740/7-14-69 (Code 471), (1973).

24. Sherby, O. D., Walser, B., Young, C. M. and Cady, E. M., "Superplastic Ultra High Carbon Steels," Third Semi-Annual Progress Report, ARPA, Grant No. DAHC-15-73-G15, (Feb. 1975).

25. Kalish, D. and Cohen, M., "Anisotropy of Properties in Martensite as Developed by Thermomechanical Treatments," Trans. ASM, 62 (1969), 353.

26. Langford, G. and Cohen, M., "Strain Hardening of Iron by Severe Plastic Deformation," Trans. ASM, 62, (1969), 623.

27. Rack, H. J., and Cohen, M., "Strain Hardening of Iron-Titanium Alloys At Very Large Strains," Mater. Sci. Eng., 6 (1970), 320.

28. Kalish, D., Kulin, S. A., and Glen, M., "Flow Strength and Fracture Toughness of 9Ni-4Co Steel as Affected by Strain Tempering and Ausforming," Metals Engr. Quart., 7, No. 4, (Nov. 1967), 54.

29. Rack, H. J. and Kalish, D., "Improved Fatigue Resistance of 18 Ni (350) Maraging Steel Through Thermomechanical Treatments," Met. Trans., 5, (March 1974), 685.

30. Brockenbrough, R. L., "Design Trends for High Strength Low Alloy Steels," Symposium on the Processing and Properties of Low Carbon Steel, J. M. Cray, editor, TMS-AIME, New York, (1973), 389.

31. Baird, J. D. and Preston, R. R., "Relationships between Processing, Structure and Properties in Low Carbon Steels," vide supra.

32. Korchynsky, M. and Stuart, H., "The Role of Strong Carbide and Sulfide Forming Elements in the Manufacture of Formable High Strength Low Alloy Steels," Symposium on Low Alloy High Strength Steel, Metallurgy Companies, Nuremburg, BRD, (1970), 17.

33. Meyer, L., Weisterkamp, F., Strassburger, C., "Alloying Possibilities for Increasing Strength and Toughness of Weldable Structural Steels," vide supra.

34. Irvine, J. J., "Strong Structural Steels," vide supra.

35. Little, J. H. Chapman, J. A., Morrison, W. B., and Mintz, B., "The Production of High Strength Plate Steels by Controlled Rolling," The Microstructure and Design of Alloys, Proc. of the 3rd Int. Conf. on the Strength of Metals and Alloys, Inst. of Metals, Cambridge, Vol. 1 (August 1973), 80-84.

36. Melloy, G. and Dennison, J. D., "Continuum Rolling--A Unique Thermomechanical Treatment for Plain-Carbon and Low Alloy Steels," vide supra, 60-64.

37. Jenkins, R. D. and Wilson, D. V., "Development of High Drawability in a Low-Carbon Rimmed Steel by Controlled Cold Rolling and Intracritical Annealing," J. Iron & Steel Inst., (Oct. 1971), 805.

38. Private communication: D. J. Walker and R. W. K. Honeycombe, "The Effect of Plastic Deformation on the Isothermal Decomposition of Metastable Austenite," presented in Cincinnati, (1975) ASM, in Press.

39. Jahnke, L. P., "The Importance of Processing Technology in the Future Development of Superalloys and the Gas Turbine," J. Metals, (April 1973), 15.

40. Shamblen, C. E., Allen, R. E., and Walker F. E., "Affect of Processing and Microstructure on Rene 95," Met. Trans. A, 6A, (Nov. 1975), 2073.

41. Wilcox, B. A. and Clauer, A. H., "The Role of Grain Size and Shape in Strengthening of Dispersion Hardened Nickel Alloys", Acta Met. 20, (May 1972), 743.

42. Ruscoe, M. J. H., Norris, L. F., Clegg, M. A., and Evans, D. J. I., "The Development of Thermomechanical Processes for Advanced Dispersion Strengthened Alloys," The Microstructure and Design of Alloys, Proc. of the Third Int. Conf. on Strength of Metals and Alloys, Inst. of Metals Cambridge, Vol. 1, (1973), 96-100.

43. Cairns, R. L., Curwick, L. R., and Benjamin, J. S., "Grain Growth in Dispersion Strengthened Superalloys by Moving Zone Heat Treatments," Met. Trans A, 6A, (Jan. 1975), 179.

44. Benjamin, J. S., "Dispersion Strengthened Superalloys by Mechanical Alloying," Met. Trans., 1, (October 1970), 2943.

45. Benjamin, J. S. and Bomford, M. J., "Effect of Yttrium Oxide Volume Fraction and Particle Size on Elevated Temperature Strength of a Dispersion Strengthened Superalloy", Met. Trans., 5, (1974), 615.

46. Benjamin, J. S. and Morse, J. P., New Trends in Materials Fabrication, to be published.

47. Private Communication from Dr. Jay Larson, INCO Laboratories.

48. Conserva, M. and Leoni, M., "Effect of Thermal and Thermo-Mechanical Processing on the Properties of Al-Mg Alloys", Met Trans. A, 6A, (Jan. 1975), 189.

49. Waldman, J., Sulinski, H., and Markus, H., "Thermomechanical Processing of Aluminum Alloy Ingots," Advanced Deformation Processing, ed. by J. J. Burke, 21st Annual Sagamore Conference, Syracuse: Syracuse University Press, (1975) in press.

50. Waldman, J., Sulinski, H., and Markus, H., "The Effect of Ingot Processing Treatments on the Grain Size and Properties of Al Alloy 7075," Met. Trans., 5, (March 1975), 573.

51. DiRusso, E., Conserva, M., Gatto, F., and Markus, H., "Thermomechanical Treatments on High Strength Al-Zn-Mg (-Cu) Alloys," Met. Trans., 4 (1973), 1133.

52. DiRusso, E., Conserva, M., and Burratti, M., "A New Thermomechanical Procedure for Improving the Ductility and Toughness of Al-Zn-Mg-Cu Alloys in the Transverse Directions," Mater. Sci. Eng., 14 (1974) 23.

53. Thompson, D. S. and Levy, S. A., "Thermomechanical Aging of Aluminum Alloys", The Microstructure and Design of Alloys, Proc. Third Int. Conf. on the Strength of Metals and Alloys, Institute of Metals, Cambridge, Vol. 1, (1973), 119-123.

54. Paton, N. E. and Sommer, A. W., "Influence of Thermomechanical Processing Treatments on Properties of Aluminum Alloys," Vide supra, 101-108.

55. Oblak, J. M. and Owczarski, W. A., "Thermomechanical Strengthening of a γ' Precipitation-Hardened Nickel-Base Alloy," Met. Trans. 3, (1972), 617.

56. Thompson, A. W. and Bernstein, I. M., Advances in Corrosion Science and Technology, 6, Plenum, New York, in press.

57. Larson, F. and Zarkades, A., "Textured Titanium Technology," Advanced Deformation Processing, ed. by J. J. Burke, 21st Annual Sagamore Conference, Syracuse: Syracuse University Press (1975), in press.

58. Amateau, M. F., Dull, D. L., and Raymond, L., "The Effect of Processing on Plastic Anisothopy of Ti-6Al-4V," Met. Trans., 5, (March 1974), 561.

59. Stubbington, C. A. and Bowen, A. W., "Improvements in the Fatigue Strength of Ti-6Al-4V Through Microstructure Control," Jrn. Mat. Sci. 9, (1974), 941.

60. Plewes, J. T., "High Strength Cu-Ni-Sn Alloys by Thermomechanical Processing", Met. Trans. A, 6A, (March 1975), 537.

FIGURE CAPTIONS

1. Schematic representation of thermomechanical treatments in relation to a TTT diagram (after May and Latham[6,7]).
2. TTT Diagram of M-50 tool steel chosen for ausforming inner races and ball bearings of jet engines (after Bamberger[14-17]). The stippled region corresponds to the temperature interval or "bay" within which the warm working was performed.
3. The effect of deformation on the rolling contact fatigue life of ausformed M-50 tool steel (after Bamberger[14-17]).
4. A comparison of the rolling contact fatigue life of ausformed and conventionally treated M-50 tool steel, in terms of B_{10}-the number of cycles for 10% failures. The ausformed steel was given a prior austenitic deformation of 80% (after Bamberger[14-17]).
5. The influence of austenite stability (with respect to strain) expressed in terms of the parameter "m" (see text) on (a) the elongation and (b) the fracture toughness, K_{IC}, for TRIP steels.
6. Yield (0.2 pct. offset) and tensile strength vs number of austenite reversions for TRIP steel (after Koppenaal[21]).
7. Schematic representation of a typical controlled rolling schedule of the type largely used in the early production of HSLA steels up to 1970 (after Little et al[35]).
8. Schematic representation of several common practices used for the TMT production of HSLA steels (after Melloy and Dennison[36]).
9. The yield strengths and transition temperatures (for a Charpy Value of 15 ft-lbs.) are shown for the steel processed by the TMT's shown in Fig. 8 (after Melloy and Dennison[36]).

10. The relation between transition temperature (for a Charpy value of 15 ft-lbs) and a parameter which describes the intensity of the cube-on-corner texture (after Melloy and Dennison[36]).

11. The TTT diagram of 1.47%Ni, 1.1%Cr, 0.27%Mo, 0.36%C steel (AISI 4337) showing the effect of different tensile deformations at 1020°F (550°C) on the onset of transformation (after Walker and Honeycombe[38]).

12. René 95 microstructures in the conventional fine grain and TMT conditions: (a) Typical fine grain microstructures (magnification 460 times); (b) at (magnification 460 times); (c) "necklace" microstructure of TMT conditions (Magnification 460 times) and (d) at (Magnification 460 times), (after Shamblen et al[40]).

13. Schematic comparison of strength as a function of aging time for conventional thermal treatment (dashed line) and for TMT (solid line). The TMT sequence was to age the material to the T6 condition, deform 15% at 465 K, and age further at 435 K. The resulting T73 SCC resistance was accompanied by T6 strength because of dislocation substructure strengthening, (after Paton and Sommer[54]).

14. The influence of texture on the mechanical properties of titanium as shown as a function of specimen orientation for: (a) Young's modulus; (b) yield strength; (c) impact energy; and (d) stress corrosion cracking, K_{ISSC} (after Larson and Zarkades[57]).

TABLE I

THERMOMECHANICAL TREATMENTS OF STEEL
(after Araki[13])

Classification of TMT		Nomenclature
I Deformation before transformation T > Md	A. - of stable γ	HTMT
	1. - to M (and/or B) 2. - to F + P (and/or B)	Ausforging (controlled rolling)
	B. - of metastable γ	LTMT (hot-cold working)
	1. - to M (and/or B)	Ausforming (Ausroll temper)
II Deformation accompanying transformation*	C. - of less stable γ	(Strain-induced transformation)
	1. - to M 2. - to F + P	Zeroll (cold working of γ) Isoforming (dynamic recovery)
III Deformation and aging	D. - of M (carbon-free)	Marforming (dynamic strain aging) (dynamic recovery)
	E. - of transformed structure	Warmworking
	1. - of M, TM 2. - of F + P, B etc.	E.T.D. etc.
	F. - of M	Warmworking at higher temperature (dynamic recovery)

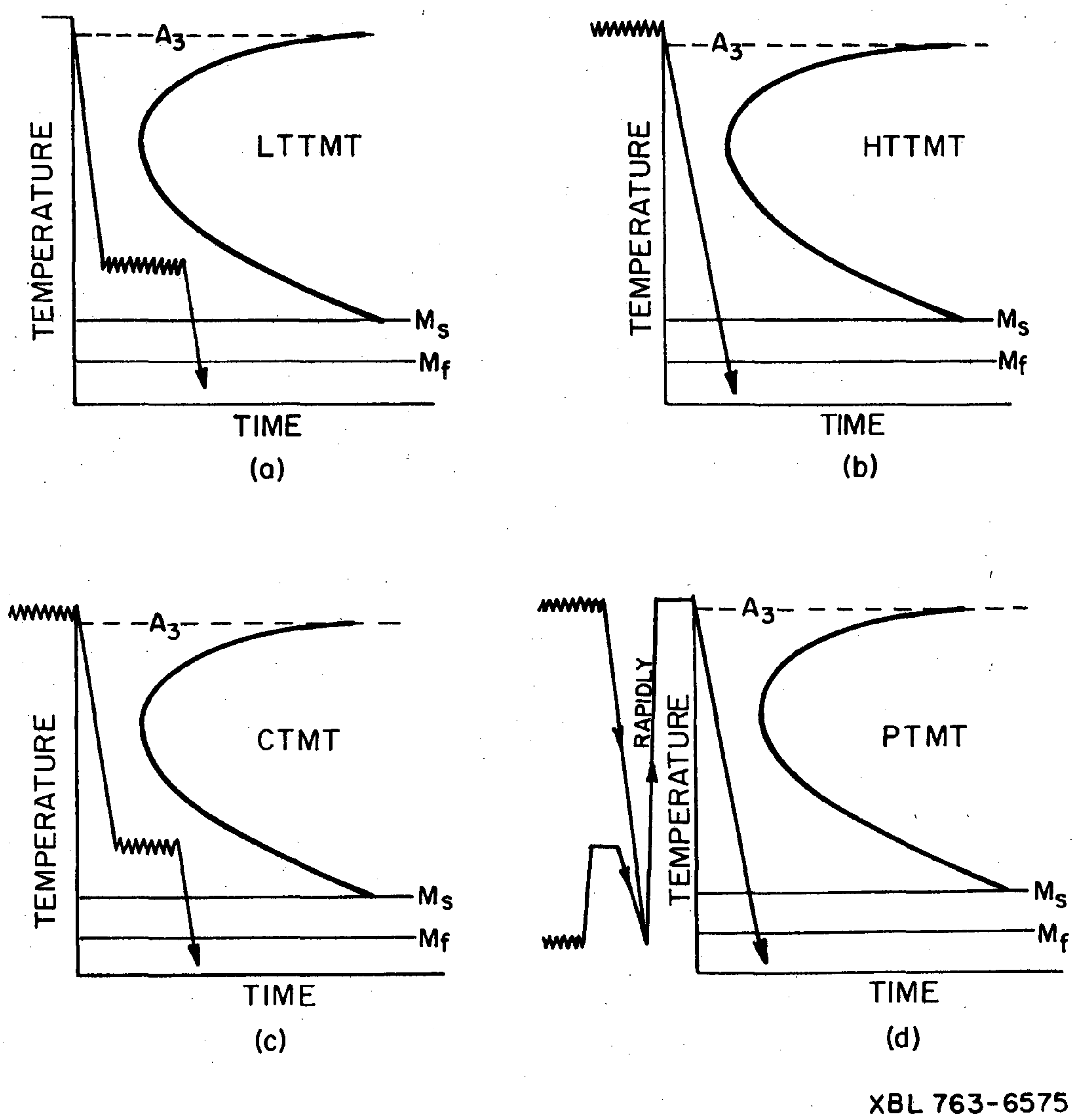
* Relevant phenomena: TRIP, etc.

TABLE II

Probably Relationships Between Stability, as Affected by Several Processing and Testing Variables, and Selected Mechanical Properties

Variable	Stability	Mechanical Property					
		Yield Strength	Luders Strain	Elongation	Work Hardening Rate	Fracture Toughness	Resistance to Hydrogen Embrittlement
COMPOSITION (increasing)							
Substitutional solutes (exception of Co)	↑ [9,11,12]	↑ [11,12]	↑ [9,11,12]	↑↓[9,11,12]	↓ [11,12]	↓ [19]	↑ [29-31]
Interstitial solutes (C and N)	↑ [9,10, 12,14]	↑ [10,12, 14]	↑ [9,10 12,14]	↑↓[9,10, 12,14]	↓ [10,12 14]	↓ [19]	↑↓[29-31]
PROCESSING (All variables increasing)							
Amount of deformation (PDA)	↓ [9]	↑ [9,12,14]	↓ [9,12,14]	↑↓[9,12-14]	↑ [12-14]	↑ [19]	↓ [29-31]
Temperature of deformation (PDA)	↓↑[11]	↑↓[11]	↓↑[11]	↑↓[11]	↑↓[11]	↑↓[19]	↓↑[29-31]
Time at temperature of deformation (PDA)	↓ [28]	↑ [28]	↓ [28]	↑↓[28]	↑ [28]	?	↓ [29-31]
TESTING							
Test Temperature (decreasing)	↓ [9-11, 14]	↑↓[9-11, 14]	↑↓[9-11, 14]	↑↓[9-11, 14]	↑↓[9,11, 14]	↑↓[15,16 19]	↑↓[29-32]
Strain Rate (increasing)	↑ [23]	↑↓[23]	↑↓[23]	↑↓[23]	↑↓[23]	↑↓[15,16 19]	↑↓[29-32

Note: [ ] refers to relevant paper in the bibliography of the orignal paper by Bhandarkar et al.[19]



A3
LTTMT
TEMPERATURE
Ms
Mf
TIME
(a)
A3
HTTMT
TEMPERATURE
Ms
Mf
TIME
(b)
A3
CTMT
TEMPERATURE
Ms
Mf
TIME
(c)
A3
RAPIDLY
TEMPERATURE
PTMT
Ms
Mf
TIME
(d)
XBL 763-6575

Figure 1

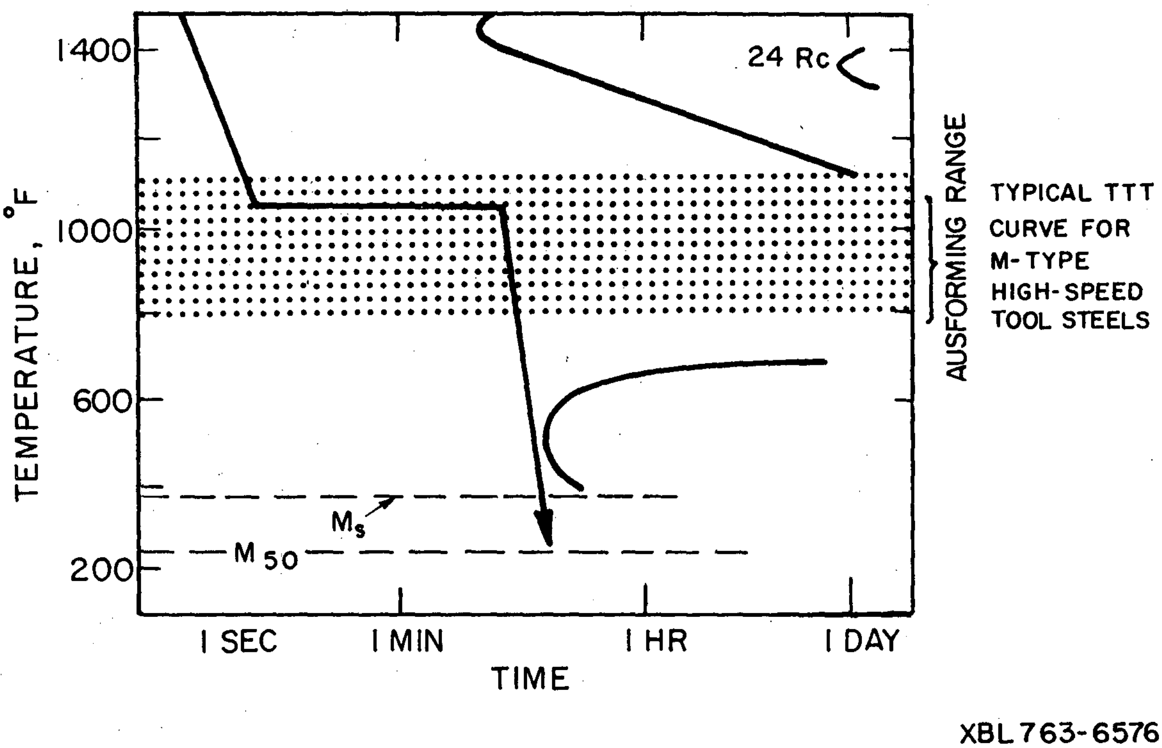
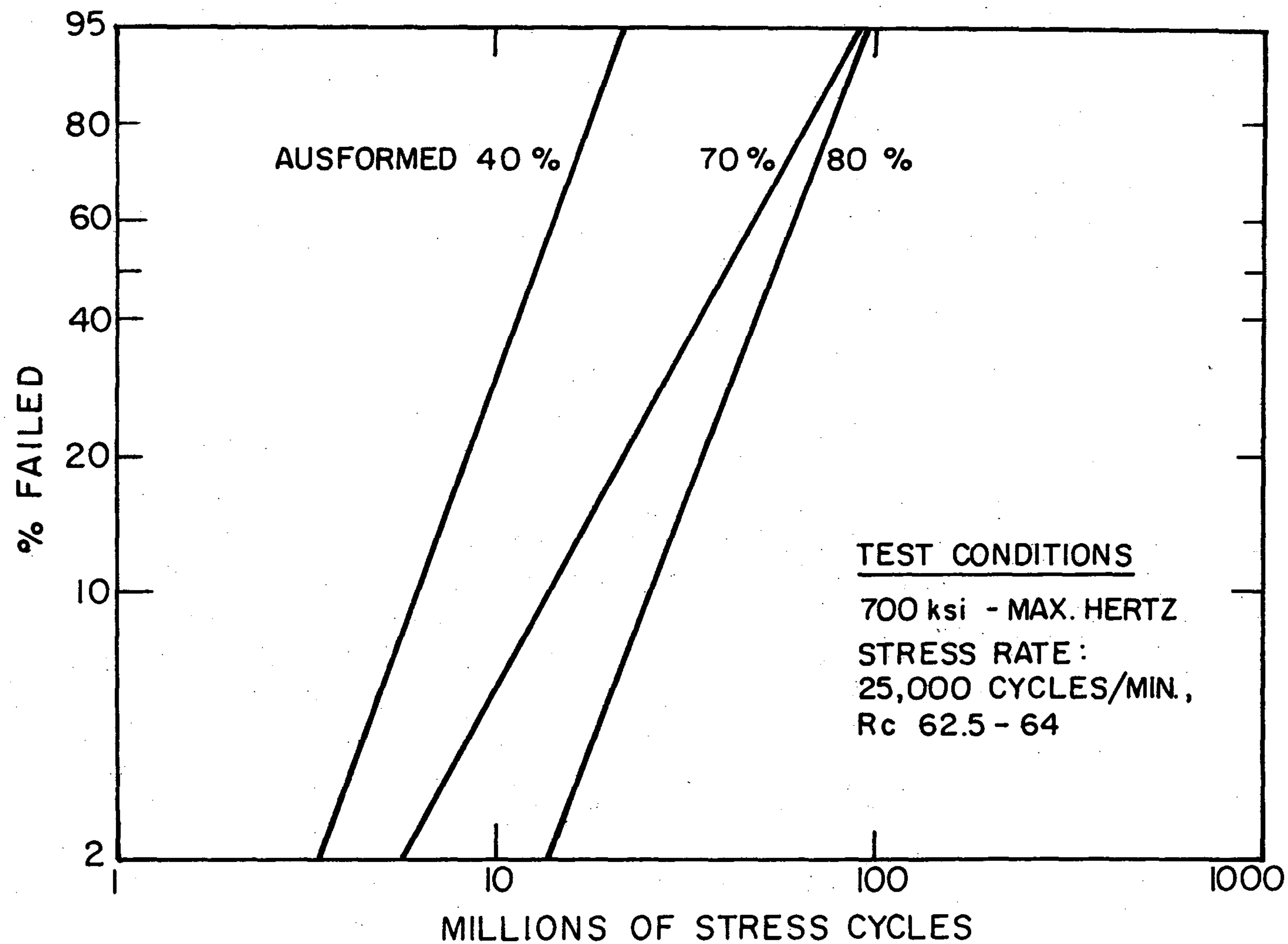

1400
1000
600
200
TEMPERATURE, °F
24 Rc
AUSFORMING RANGE
TYPICAL TTT
CURVE FOR
M-TYPE
HIGH-SPEED
TOOL STEELS
Ms
M 50
I SEC
I MIN
I HR
I DAY
TIME
XBL 763-6576

Figure 2



XBL763-6577

Figure 3

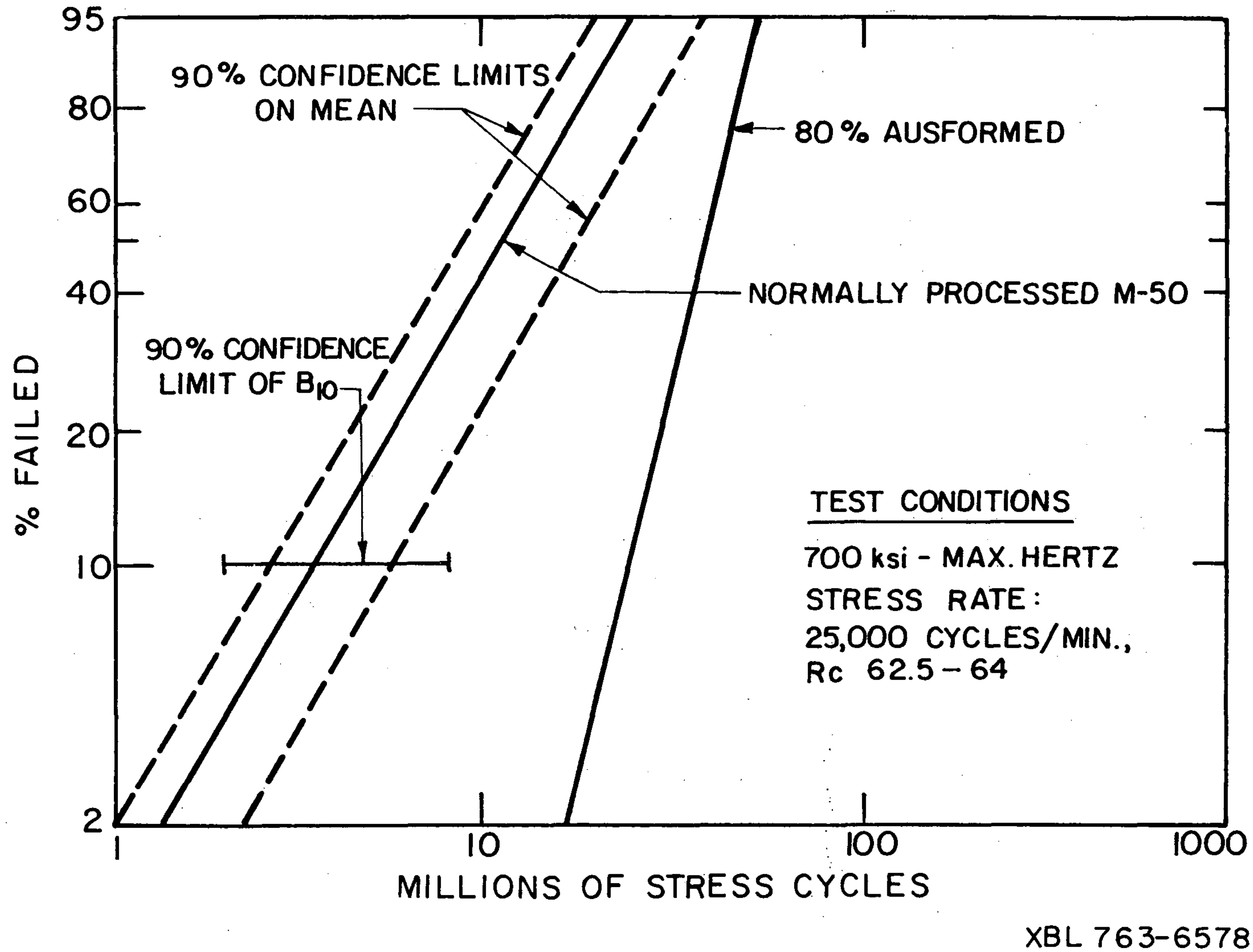

90% CONFIDENCE LIMITS ON MEAN
80% AUSFORMED
NORMALLY PROCESSED M-50
90% CONFIDENCE LIMIT OF B_{10}
TEST CONDITIONS
700 ksi - MAX. HERTZ
STRESS RATE:
25,000 CYCLES/MIN.,
Rc 62.5 - 64
% FAILED
95
80
60
40
20
10
2
1
10
100
1000
MILLIONS OF STRESS CYCLES
XBL 763-6578

Figure 4

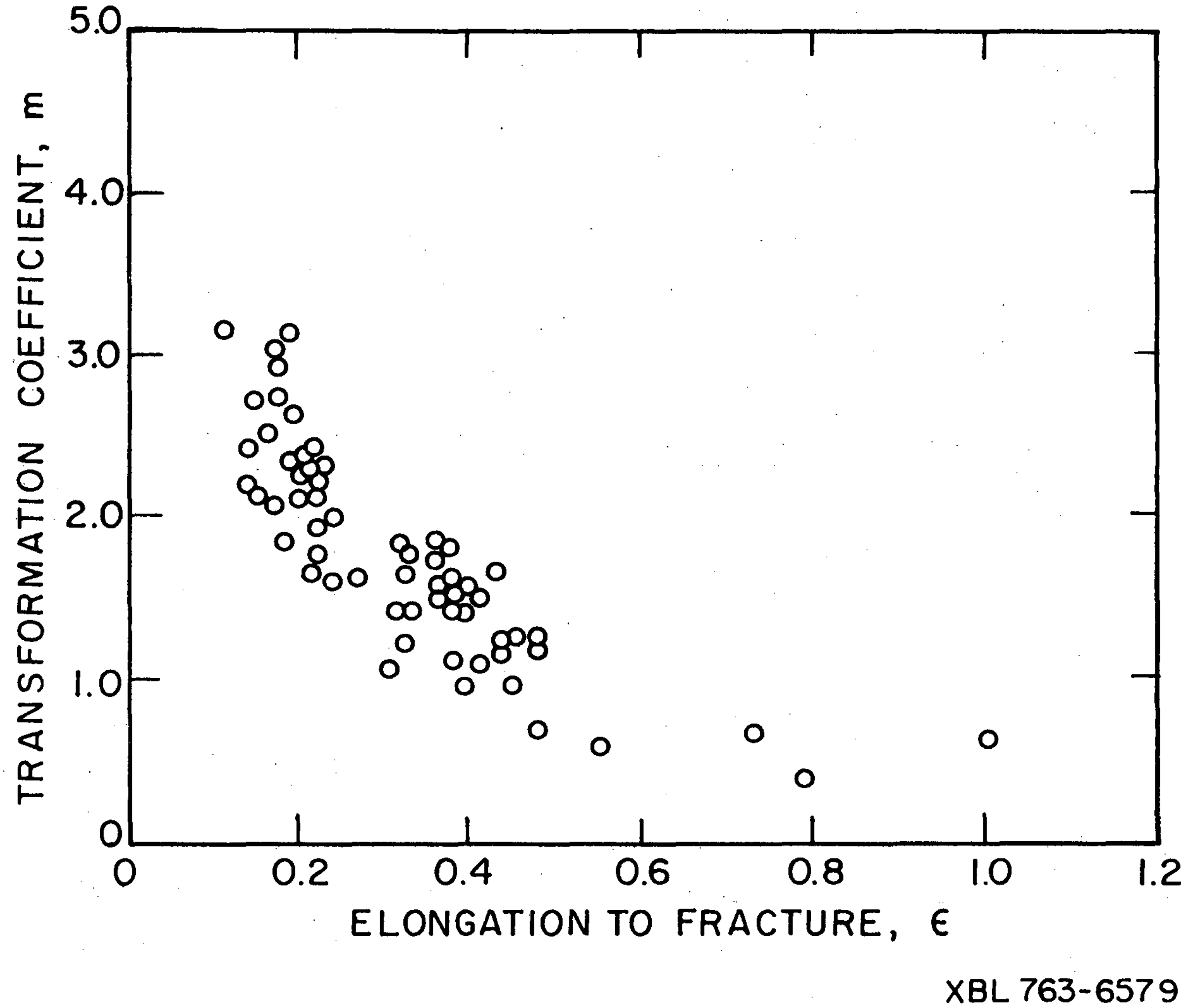

TRANSFORMATION COEFFICIENT, m
5.0
4.0
3.0
2.0
1.0
0
0
0.2
0.4
0.6
0.8
1.0
1.2
ELONGATION TO FRACTURE, ϵ
XBL 763-6579

Figure 5(a)

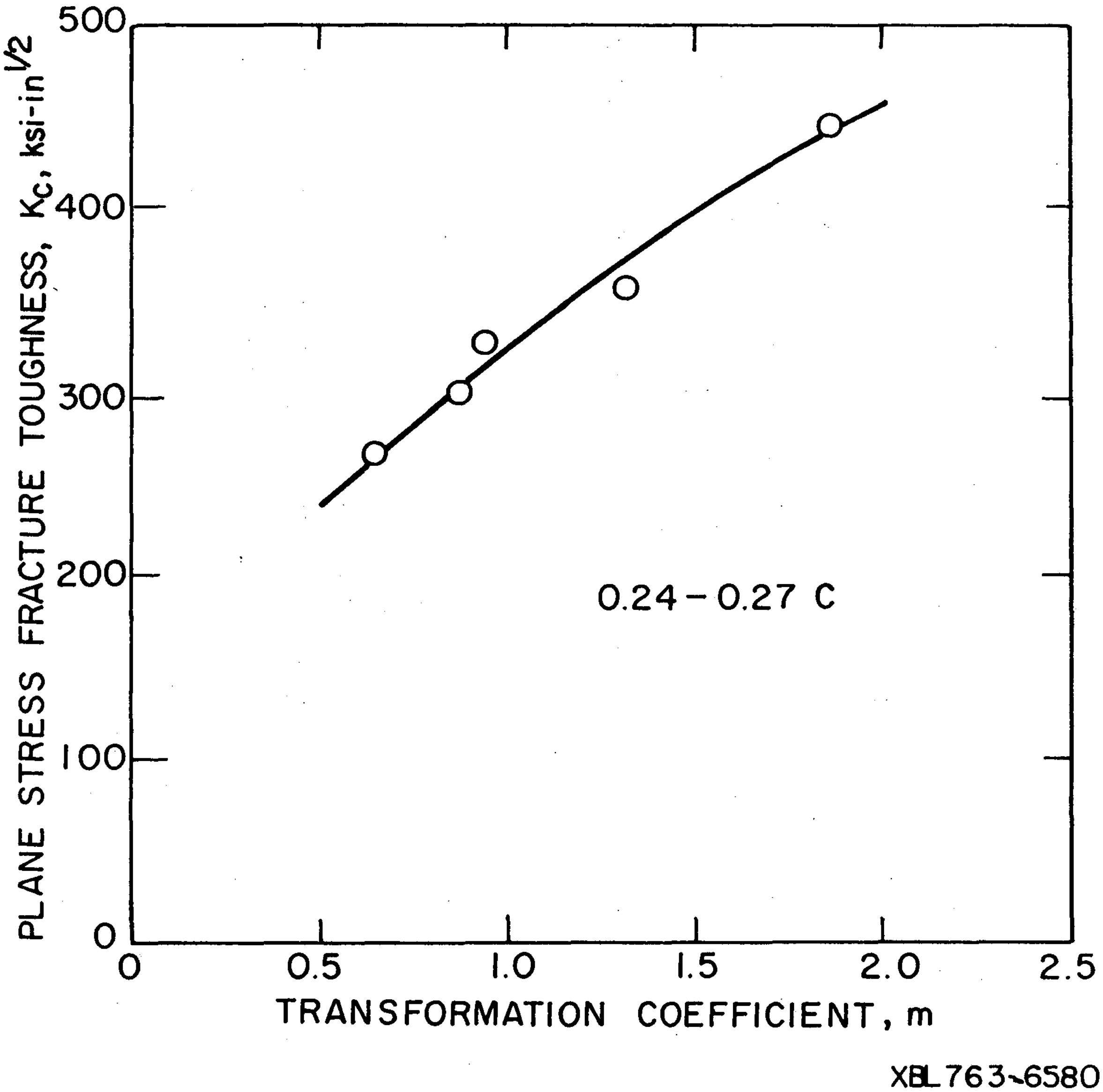

PLANE STRESS FRACTURE TOUGHNESS, K_c, ksi-in$^{1/2}$
500
400
300
200
100
0
0
0.5
1.0
1.5
2.0
2.5
TRANSFORMATION COEFFICIENT, m
0.24 - 0.27 C
XBL 763-6580

Figure 5(b)

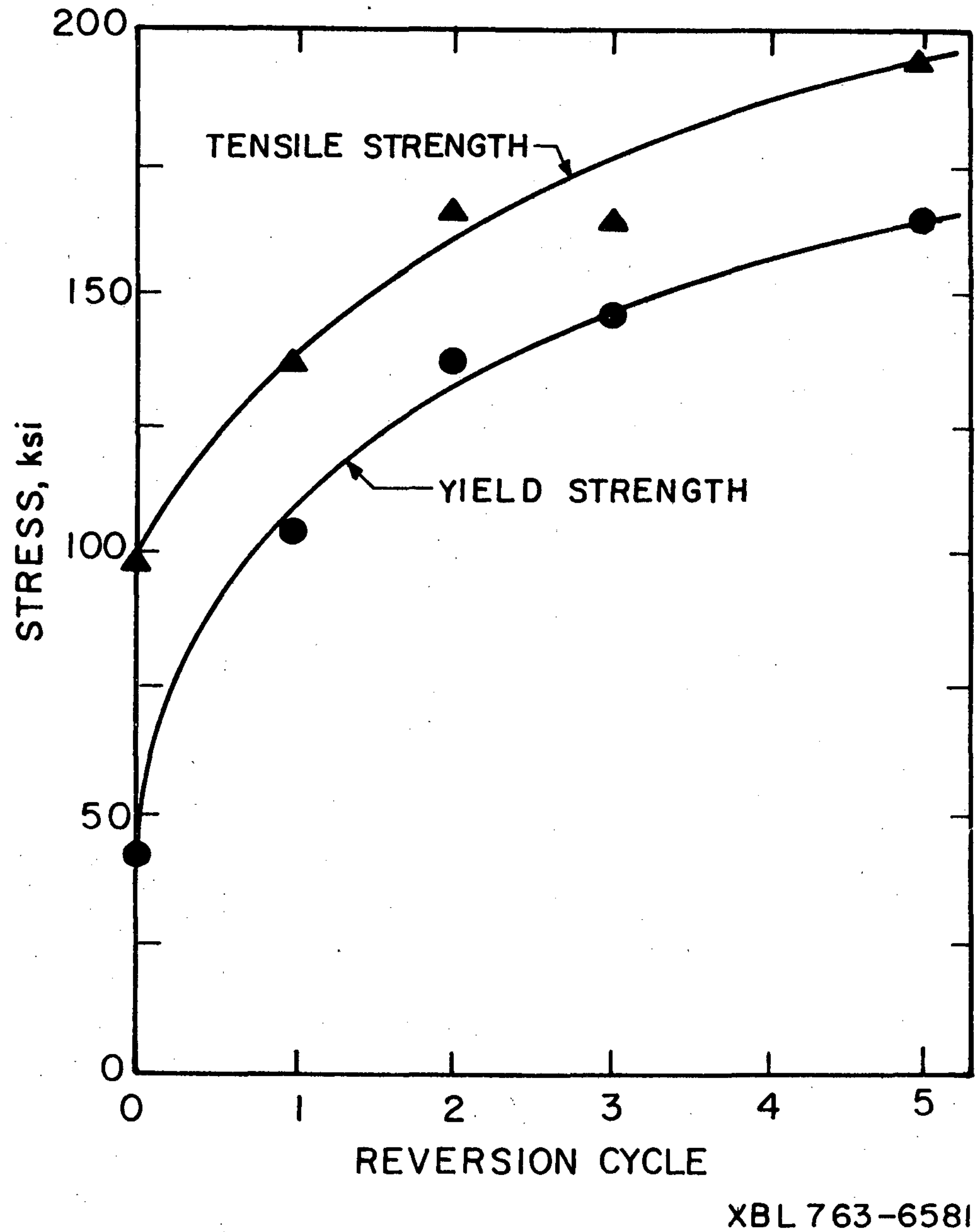

200
150
100
50
0
STRESS, ksi
TENSILE STRENGTH
YIELD STRENGTH
0
1
2
3
4
5
REVERSION CYCLE
XBL 763-6581

Figure 6

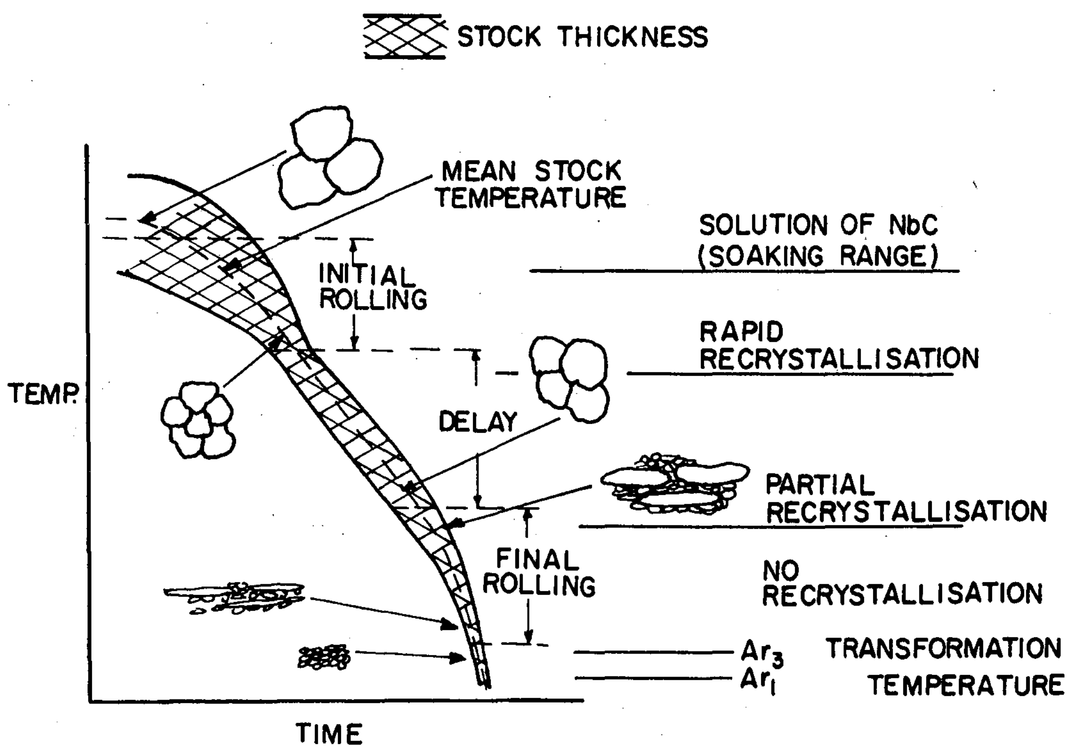
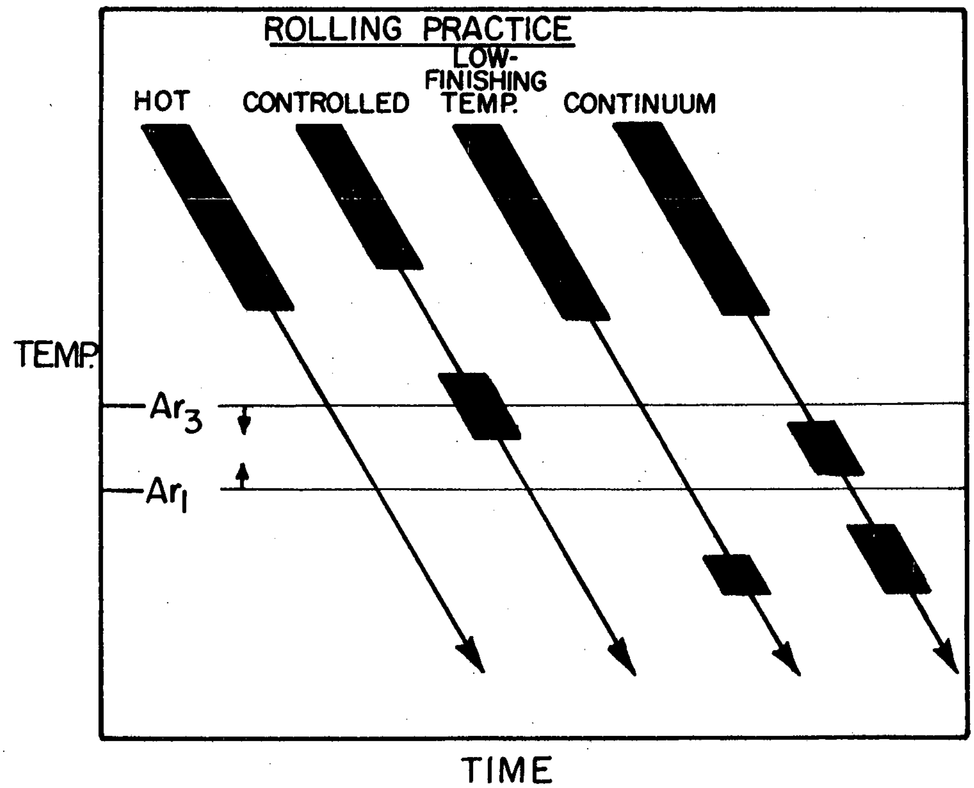
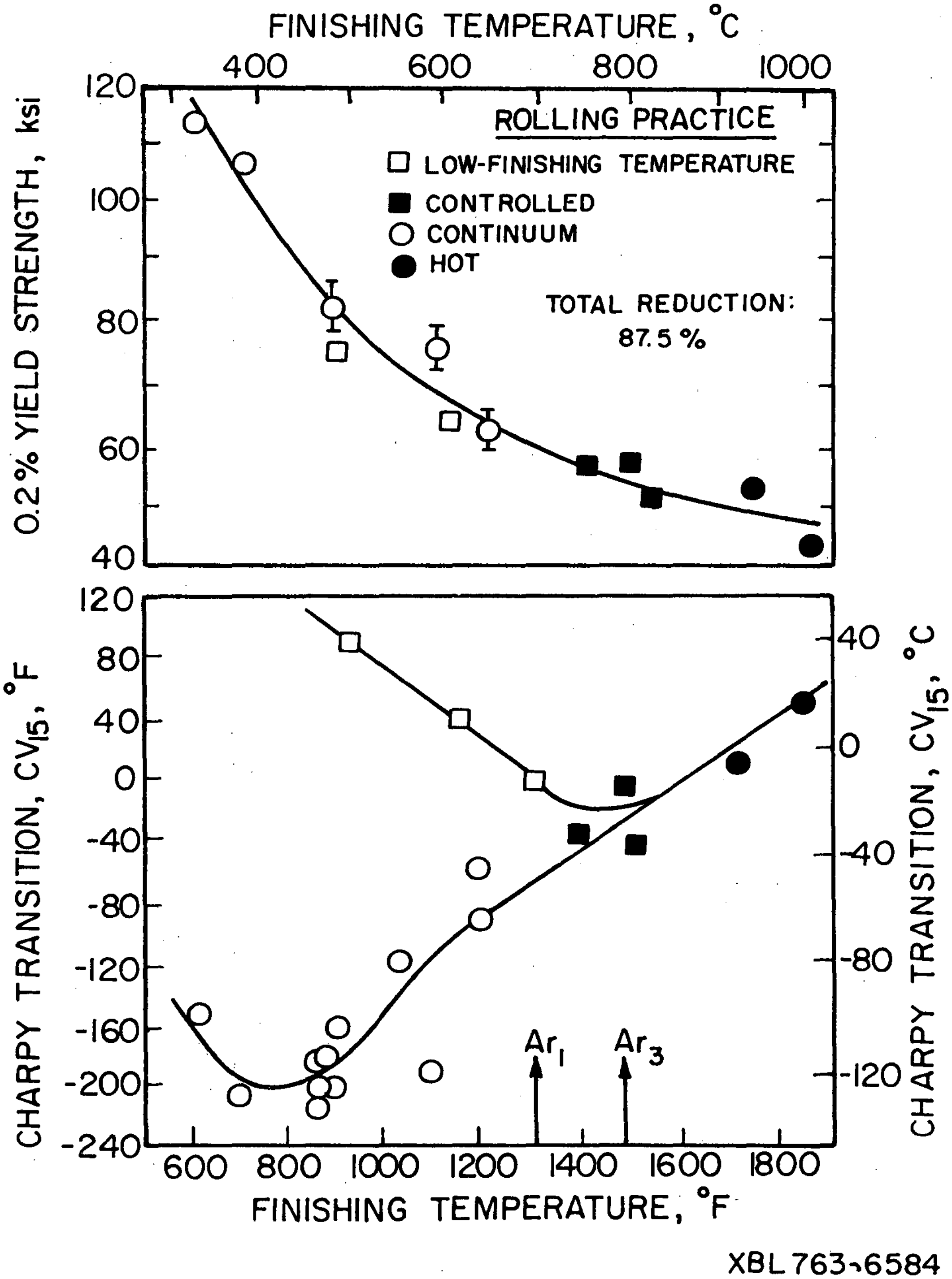

STOCK THICKNESS
MEAN STOCK TEMPERATURE
SOLUTION OF NbC (SOAKING RANGE)
INITIAL ROLLING
RAPID RECRYSTALLISATION
TEMP.
DELAY
PARTIAL RECRYSTALLISATION
FINAL ROLLING
NO RECRYSTALLISATION
Ar_3
Ar_1
TRANSFORMATION TEMPERATURE
TIME
XBL 763-6582

Figure 7



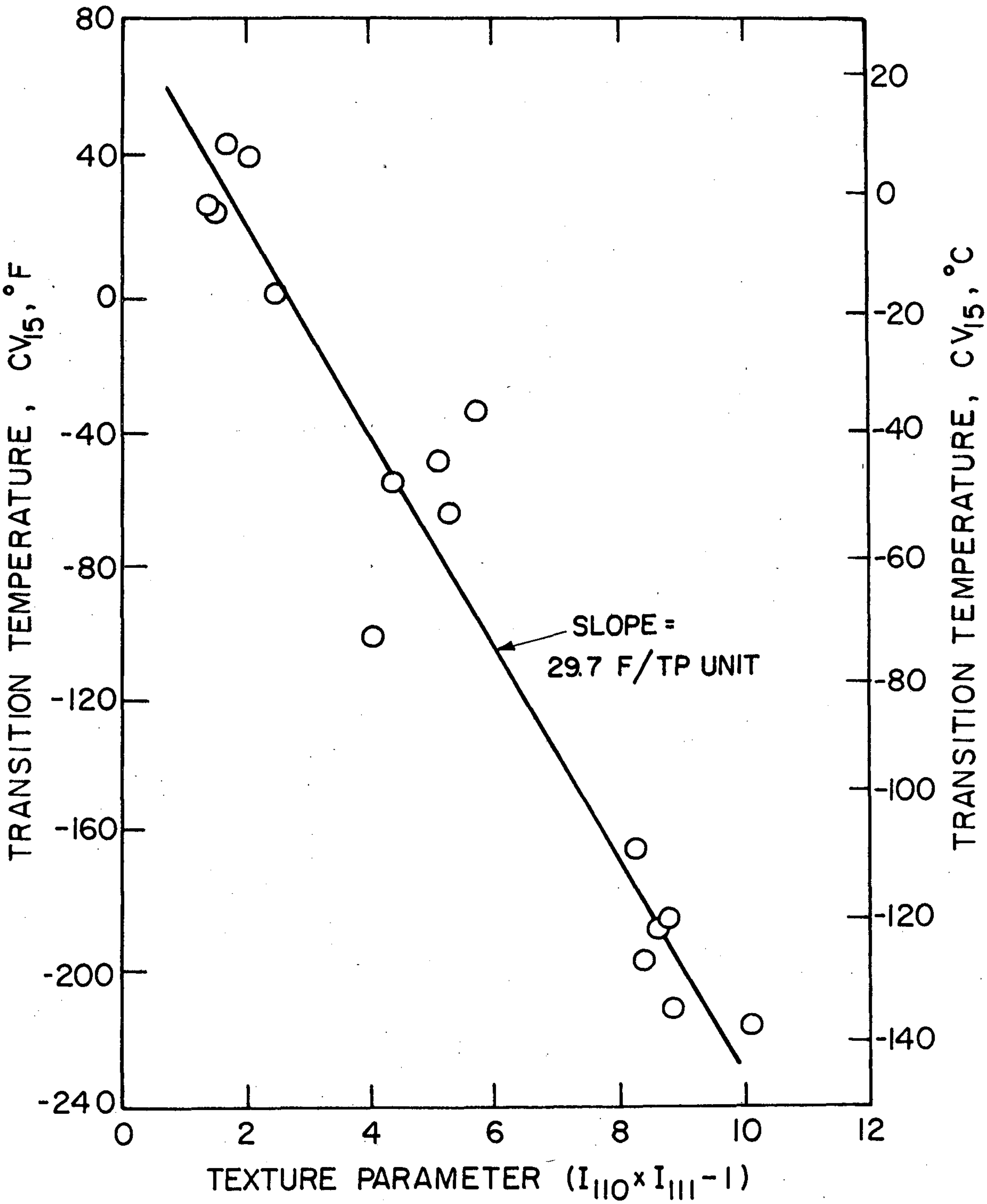
XBL 763-6583

Figure 8



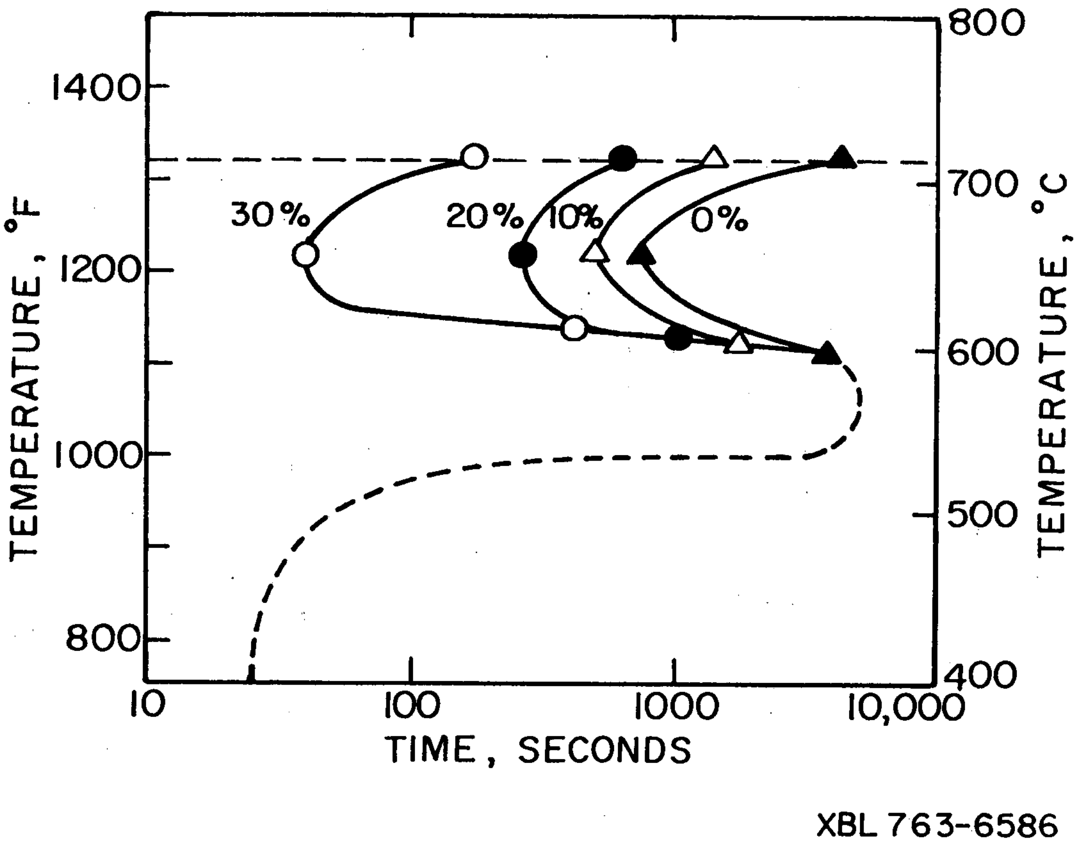
FINISHING TEMPERATURE, °C
400
600
800
1000
120
100
80
60
40
0.2% YIELD STRENGTH, ksi
ROLLING PRACTICE
LOW-FINISHING TEMPERATURE
CONTROLLED
CONTINUUM
HOT
TOTAL REDUCTION:
87.5 %
120
80
40
0
-40
-80
-120
-160
-200
-240
CHARPY TRANSITION, CV15, °F
40
0
-40
-80
-120
CHARPY TRANSITION, CV15, °C
Ar1
Ar3
600
800
1000
1200
1400
1600
1800
FINISHING TEMPERATURE, °F
XBL763-6584

Figure 9



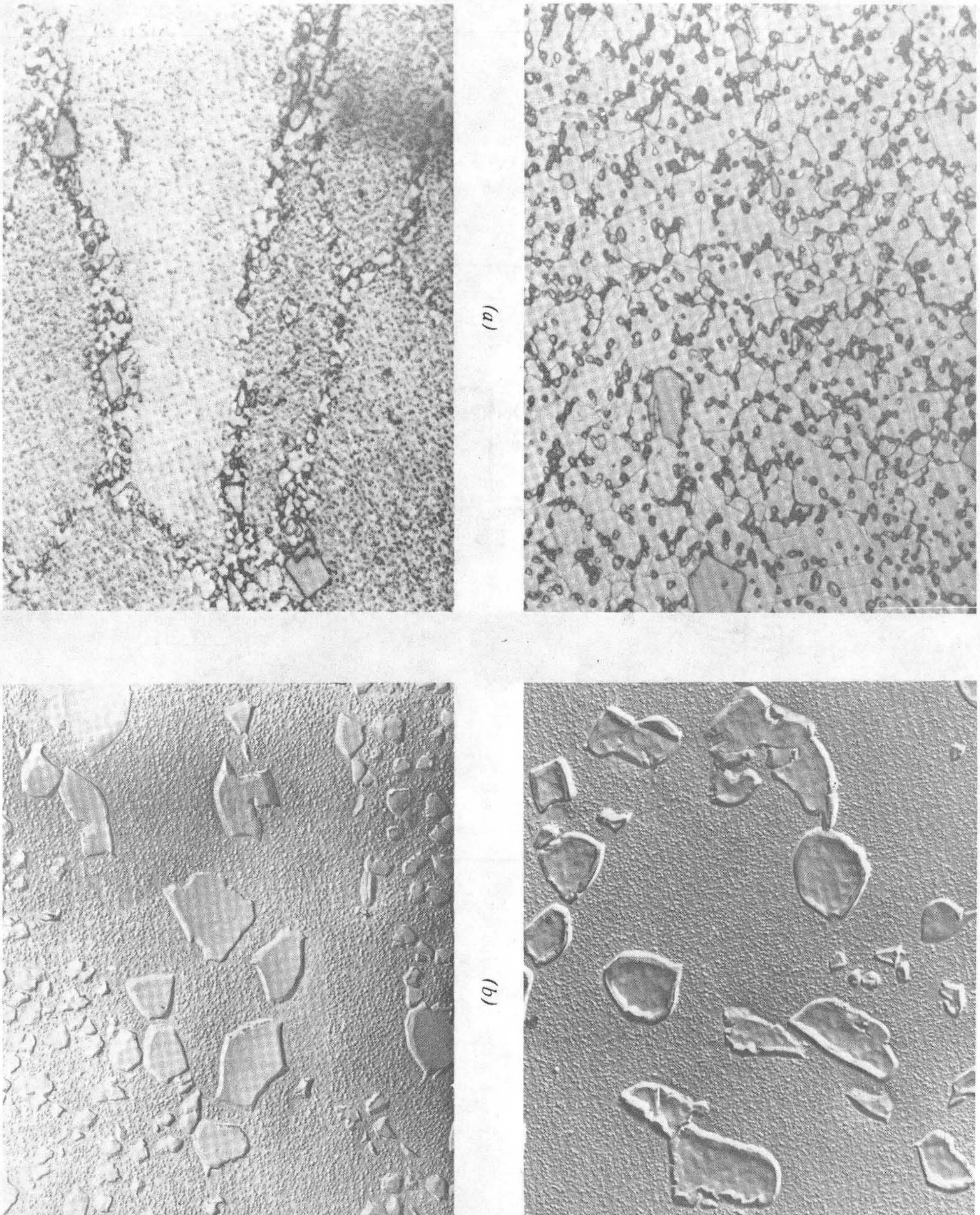
XBL 763-6585

Figure 10



1400
1200
1000
800
TEMPERATURE, °F
800
700
600
500
400
TEMPERATURE, °C
30%
20%
10%
0%
10
100
1000
10,000
TIME, SECONDS
XBL 763-6586

Figure 11



XBB 763-2145

Figure 12

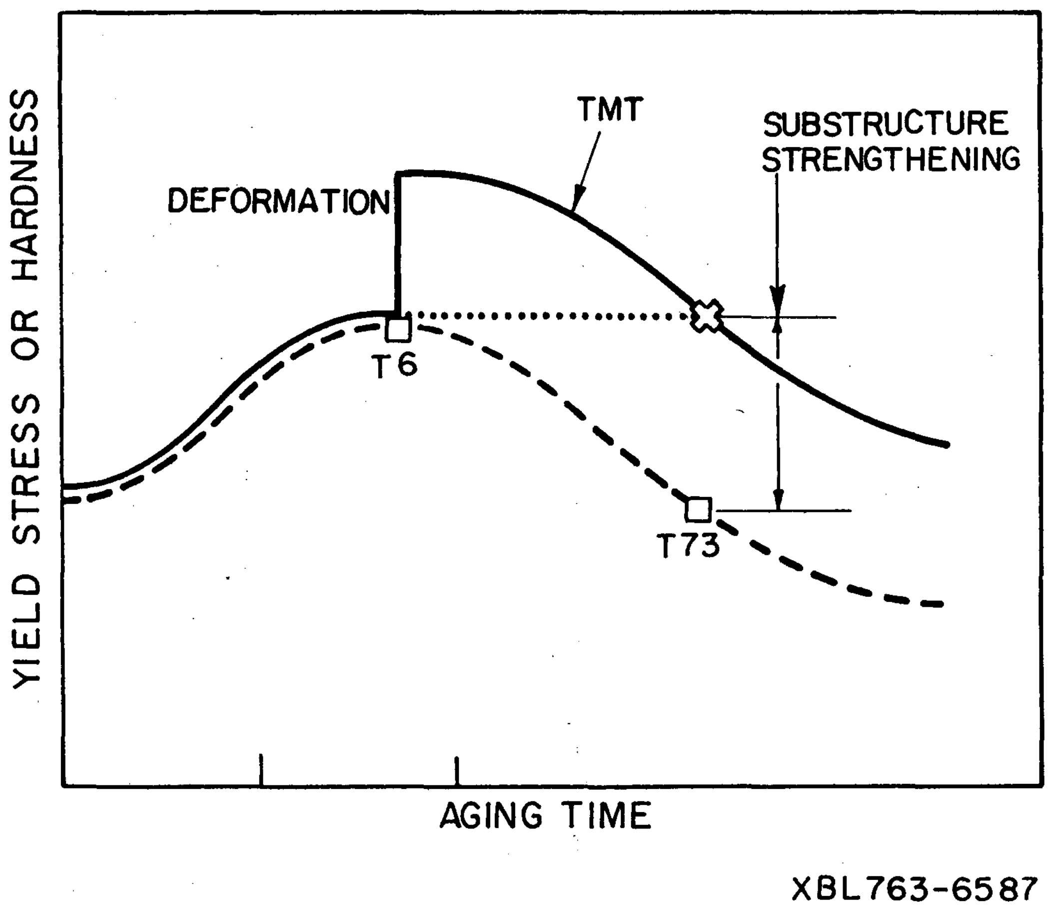
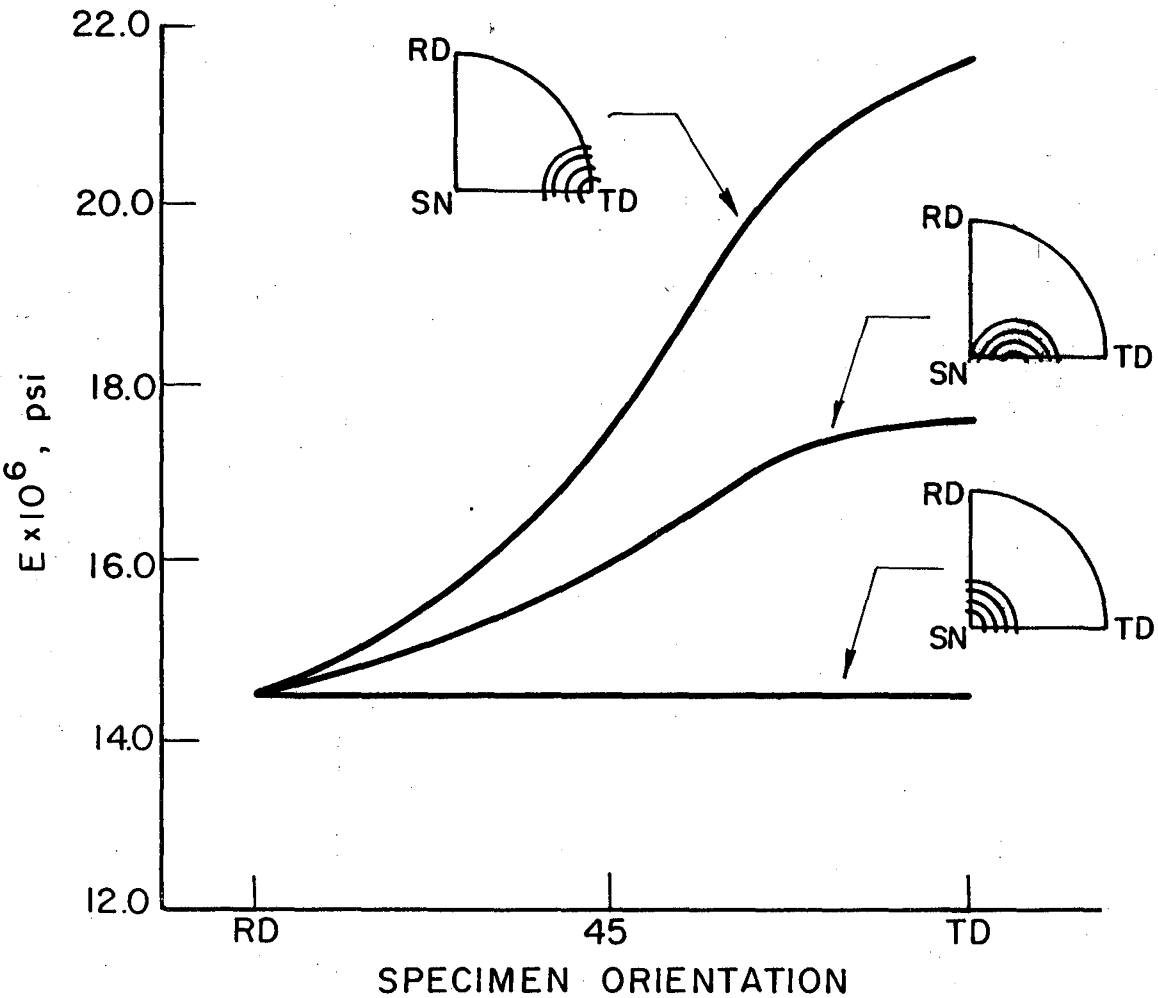

TMT
SUBSTRUCTURE
STRENGTHENING
DEFORMATION
T6
T73
YIELD STRESS OR HARDNESS
AGING TIME
XBL763-6587

Figure 13



XBL 763-6588

Figure 14(a)

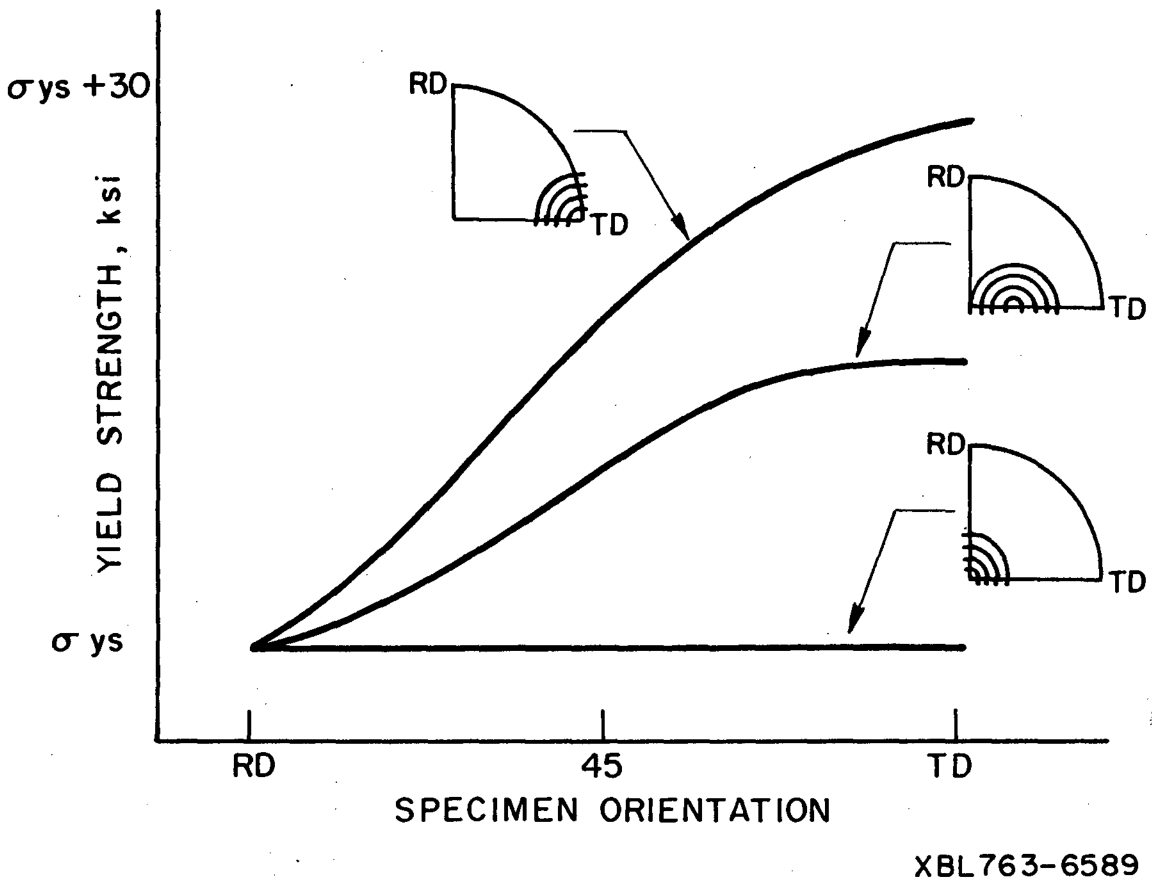

σys +30
σ ys
YIELD STRENGTH, ksi
RD
TD
RD
TD
RD
TD
RD
45
TD
SPECIMEN ORIENTATION
XBL763-6589

Figure 14(b)

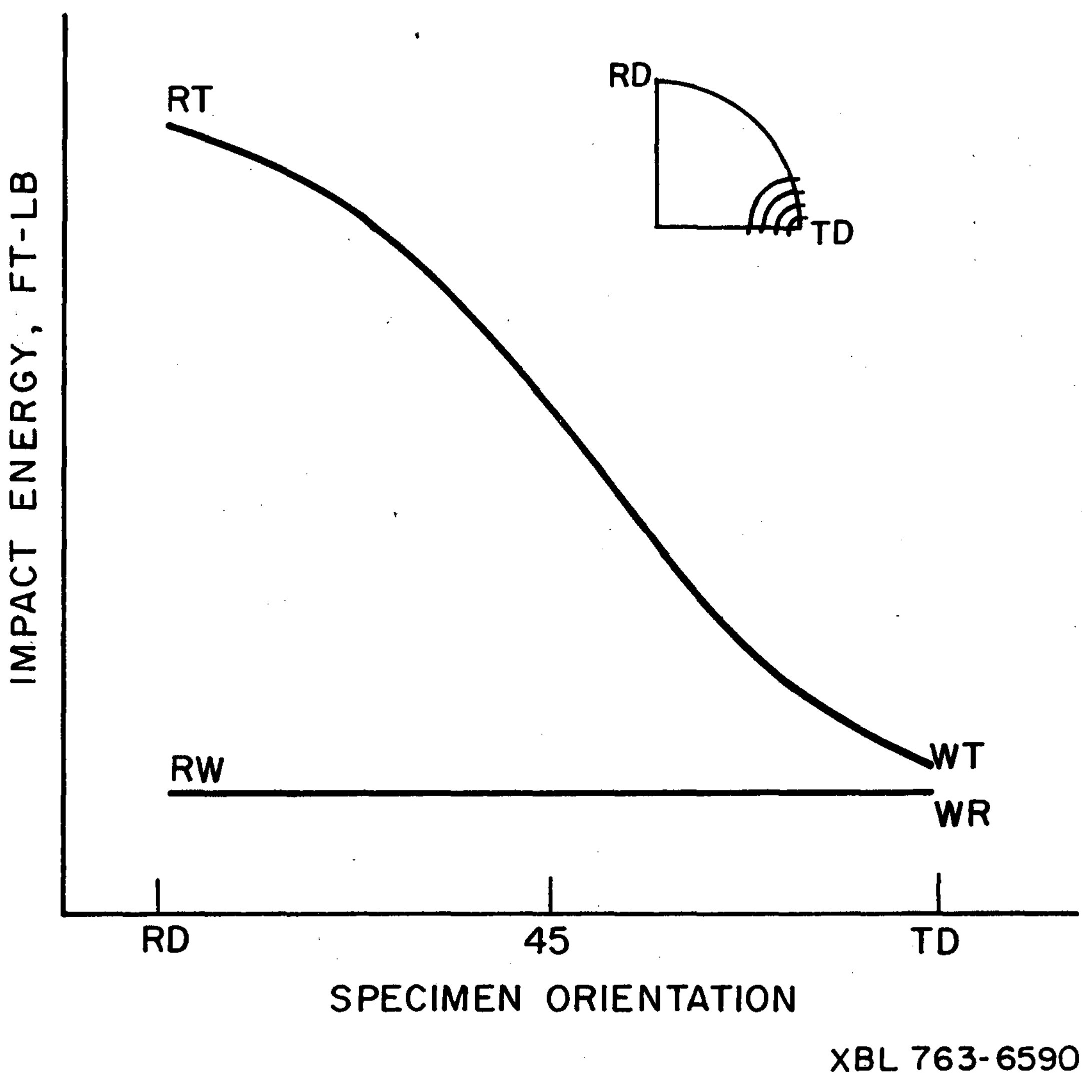

RD
TD
RT
WT
RW
WR
IMPACT ENERGY, FT-LB
RD
45
TD
SPECIMEN ORIENTATION
XBL 763-6590

Figure 14(c)

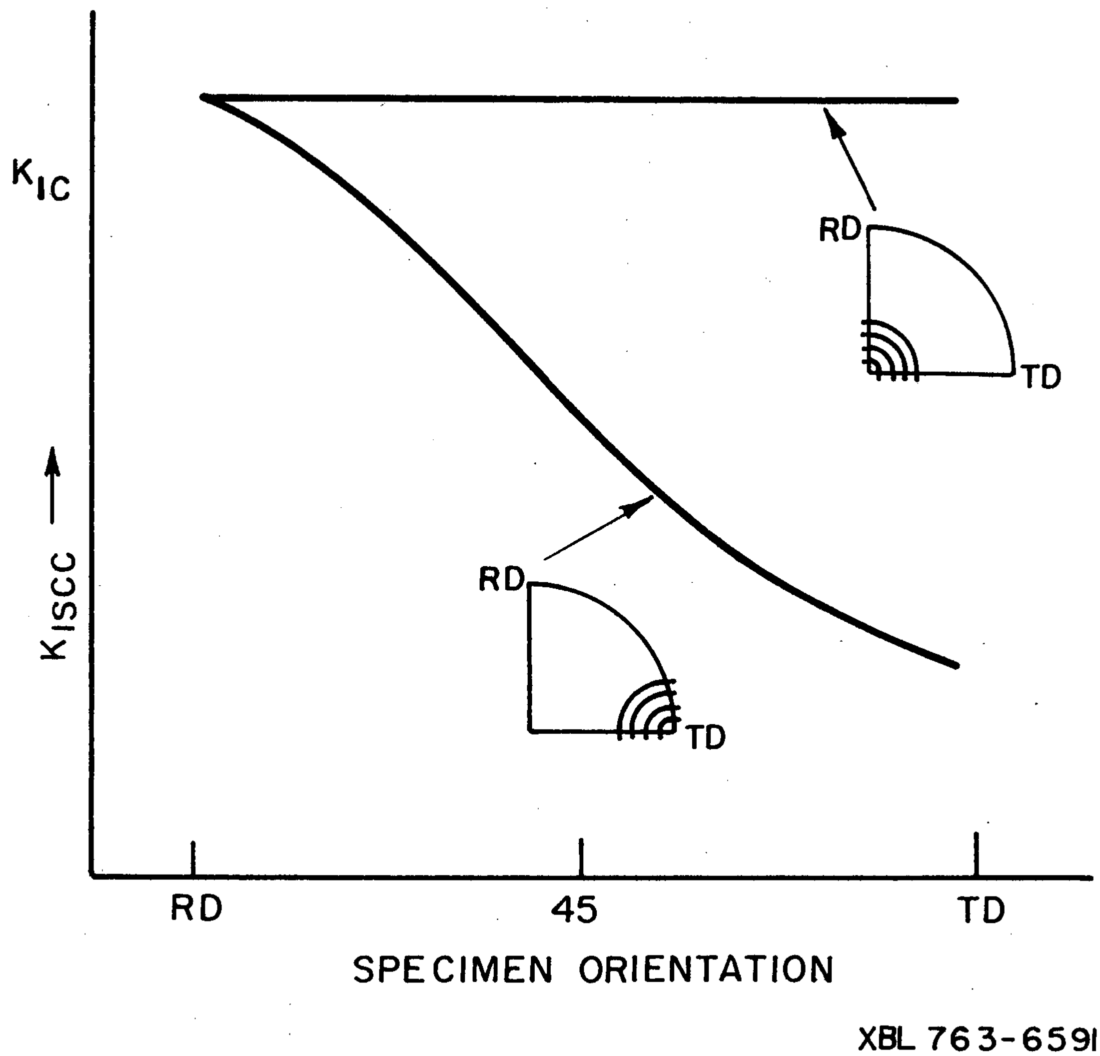

K_{IC}
K_{ISCC}
RD
TD
RD
TD
RD
45
TD
SPECIMEN ORIENTATION
XBL 763-6591

Figure 14(c)

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