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THERMAL AND MECHANICAL TREATMENT FOR PRECIPITATION-HARDENING STAINLESS STEELS

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THERMAL AND MECHANICAL TREATMENT FOR PRECIPITATION-HARDENING STAINLESS STEELS

By

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ABSTRACT

The thermal and mechanical treatments employed in the production and fabrication of precipitation-hardenable stainless steels are discussed. Particular emphasis is directed to the interaction of the processing variables and their effects on fabricability and mechanical properties. Equipment and procedures are described as well as problems likely to be encountered and precautions to be observed in processing. Detailed processing information is presented for the precipitation-hardenable stainless steels considered to be applicable for missile and aerospace applications.

FOREWORD

Precipitation-hardening stainless steels are potentially useful wherever corrosion resistance and high strength at high temperatures are needed. They were developed initially to meet urgent requirements in World War II, but new alloys and methods of processing have since been introduced to assist engineers concerned with missiles and space vehicles and with various applications in the field of nuclear science and technology.

The Atomic Energy Commission and National Aeronautics and Space Administration have established a cooperative program to make available information, describing the technology resulting from their research and development efforts, which may have commercial application in American industry. This publication is one of the many resulting from the cooperative effort of these agencies to transfer technology to private industry.

This survey is based on information contained in a series of reports originally prepared by Battelle Memorial Institute for the Manufacturing Engineering Laboratory of the George C. Marshall Space Flight Center. The original information has been updated and revised in writing the current, seven volume survey. These volumes were prepared under a contract with the NASA Office of Technology Utilization which was monitored by the Redstone Scientific Information Center.

PREFACE

This report is one of a series of state-of-the-art reports being prepared by Battelle Memorial Institute, Columbus, Ohio, under Contract No. DA-01-021-AMC-11651(Z), in the general field of materials fabrication.

This report is intended to provide information useful to heat treaters, fabricators and process engineers responsible for the processing of precipitation-hardenable stainless steels. The subjects covered in the report are physical metallurgy, annealing, solution treating, homogenizing, austenite conditioning, precipitation hardening, hot working and cold working. Heat treating and cleaning equipment are described and precautions regarding furnace atmospheres and cleaning solutions are discussed. Processing information is presented for a number of commercially available and promising developmental precipitation-hardenable stainless steels.

Information for this report was obtained from the technical literature, reports on Government contracts, producer's data, and personal contacts with technical personnel in industry. A total of 42 references are cited; most of them dated later than 1963. The authors would like to acknowledge the efforts of Mrs. Anita L. Coles, Mr. W. L. Buckel, Mr. J. A. Gibson, and Mr. V. W. Ellzey for their assistance in the literature searches and for providing the technical reports.

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THERMAL AND MECHANICAL TREATMENT FOR PRECIPITATION-HARDENING STAINLESS STEELS

SUMMARY

Precipitation-hardenable stainless steels are subjected to a variety of thermal and mechanical treatments during their fabrication into usable hardware. Among these are: hot-working operations, such as hot forging or hot rolling to produce a mill product or a raw product form; cold working such as bending, deep drawing, stretching or spinning to produce a desired final shape; and, thermal treatments such as annealing, homogenizing, solution annealing, austenite conditioning and precipitation hardening to condition the metal for fabrication or to develop in the material the microstructure and properties desired. The manner in which the material is affected depends on the type of alloy and the conditions of the specific thermal or mechanical treatment.

This report contains a brief discussion of the physical metallurgy of the PH stainless steels to illustrate how chemical composition establishes the alloy type (whether martensitic, semiaustenitic or austenitic) and to provide background information for an understanding of the metallurgical structures developed by thermal and mechanical processing. The sections on thermal treatments and mechanical treatments comprise discussions, in general terms, of the processes used and their effects on the structure and properties of the three types of precipitation-hardenable stainless steels. Information also is presented on the

equipment and procedures for heat treatment and descaling as well as on precautions to be observed regarding the effects of furnace atmospheres and cleaning solutions.

In the final sections of the report the specific effects of various thermal and mechanical treatments on mechanical properties are presented for each alloy, chiefly in the form of tables and figures. The metallurgical structures developed by the treatments are discussed and correlated with the properties in each case.

The alloys selected include currently used and newly developed materials of particular interest for missile and aerospace application. The specific steels grouped by alloy type are:

Martensitic

- 17-4 PH
- 15-5 PH
- PH 13-8 Mo
- AM 362
- AM 363
- Custom 455
- AFC-77

Semiaustenitic

- 17-7 PH
- PH 15-7 Mo
- PH 14-8 Mo
- AM 350
- AM 355

Austenitic

- A 286

INTRODUCTION

During World War II, the need for high-strength, corrosion-resistant materials of reasonable cost, that would retain considerable strength up to moderately elevated temperatures, led to the development of the precipitation-hardenable stainless steels. The first of these alloys was introduced in 1946. In the succeeding twenty years, a number of stainless steels hardened by precipitation heat treatment have been widely used in the manufacture of high performance aircraft, missiles and other military and civilian articles and structures.

Three types of precipitation-hardenable stainless steels are presently available. All are essentially austenitic at their normal annealing temperatures and are classified according to their austenite stability. The three types are martensitic, semiaustenitic, and austenitic. The compositions of currently used and newly developed precipitation-hardenable stainless steels are listed in Table I.

The martensitic precipitation hardenable alloys are 17-4 PH, 15-5 PH, PH 13-8 Mo, AM 362, AM 363, AFC-77 and Custom 455. These materials are normally used as bar or forgings, although some are available as castings or sheet and plate. The cold forming of complex parts from sheet product is difficult because the untempered martensitic structure developed on annealing has relatively high strength and low ductility. Hardening by a single low-temperature precipitation treatment will produce yield strengths of from 170,000 to 200,000 psi. The martensitic alloys can be used to temperatures as high as 900 F.

TABLE I. PRECIPITATION-HARDENABLE STAINLESS STEELS

Alloy Type and Designation	Originator	Chemical Composition, percent[a]											
		C	Cr	Ni	Cu	Al	Ti	Mo	Co	Cb + Ta	Si	Mn	Other
Martensitic													
17-4PH	Armco	0.07 max	15.5-17.5	3-5	3-5					0.15-0.45	1 max	1 max	
15-5PH	Armco	0.07 max	14.0-15.5	3.5-5.5	2.5-4.5					0.15-0.45	1 max	1 max	
PH13-8Mo	Armco	0.06 max	12.0-13.5	7-9		0.80-1.20		1.75-2.50			0.50 max	0.50 max	
AM362	Allegheny	0.03	14.5	6.5			0.80				0.20	0.30	
AM363	Allegheny	0.04	11.5	4.25			0.50				0.15	0.20	
Custom 455	Carpenter	0.03 max	11-13	7-10	1-3		0.90-1.40			0.25-0.50	0.50 max	0.50 max	B, 0.005 max
AFC-77	Crucible	0.15	14.5					5	13.5				V, 0.5
Semiaustenitic													
17-7PH	Armco	0.09 max	16-18	6.50-7.75		0.75-1.50					1 max	1 max	
PH15-7Mo	Armco	0.09 max	14-16	6.50-7.75		0.75-1.50		2-3			1 max	1 max	
PH14-8Mo	Armco	0.05 max	13.5-15.5	7.5-9.5		0.75-1.50		2-3			1 max	1 max	
AM350	Allegheny	0.08	16.50	4.30				2.75			0.25	0.80	N, 0.10
AM355	Allegheny	0.13	15.50	4.30				2.75			0.25	0.95	N, 0.10
Austenitic													
A286	Allegheny	0.08 max	13.5-16	24-27		0.35 max	1.9-2.3	1.0-1.75			0.4-1.0	1-2	B, 0.003-0.01 V, 0.1-0.5

(a) Composition of castings may be altered slightly from nominal.

The semiaustenitic precipitation-hardenable stainless steel include 17-7 PH, PH 15-7 Mo, PH 14-8 Mo, AM 350, and AM 355. These alloys are produced primarily as sheet because the austenitic structure obtained on annealing provides excellent formability. The transformation to martensite prior to precipitation hardening can be accomplished by mechanical deformation or by thermal processes. Yield strengths over 200,000 psi can be obtained with some semiaustenitic compositions and usable strength is retained to temperatures as high as 900 F.

The Alloy A-286 is a typical example of austenitic precipitation-hardenable stainless steels. These alloys have lower room-temperature mechanical properties, around 100,000 psi yield strength, but retain a large proportion of this strength to temperatures as high as 1300 F.

A wide range of engineering properties may be obtained through variations in the processing of the precipitation-hardenable stainless steels. The purpose of this report is to present current information on the thermal and mechanical treatments used for precipitation-hardenable stainless steels of each type, outline the processing methods, give precautions to be observed in processing, and show the effects of variations in treatment on the mechanical properties of the alloys.

The mechanical properties of a precipitation-hardenable stainless steel are determined by its metallurgical structure. In turn, the metallurgical structure of the steel, as a piece of finished hardware, is the result of all the processing used to manufacture the part. Therefore, the first section of the report is designed to provide a basic understanding of how processing variables can affect the metallurgical structure. This section is devoted to the physical metallurgy of the precipitation-hardenable stainless steels.

Subsequent sections describe the thermal treatments and the mechanical treatments normally used to process the precipitation-hardenable stainless steels. The purposes of the treatments are explained, terms are defined, equipment and procedures are described, and precautions and corrective measures for common problems are discussed. The final section presents mechanical-property data for each type of precipitation-hardenable stainless steel, and the effects of variables in the thermal and mechanical treatments on the properties obtained. These data are shown to illustrate the effects of processing variables on properties and should not be considered as minimum or design values.

PHYSICAL METALLURGY

Physical metallurgy is the study of the physical and mechanical properties of metals as they are affected by composition, heat treatment, and mechanical working. A brief discussion of the physical metallurgy of stainless steel is presented here to provide background necessary for a better understanding of the effects of processing parts and articles in production. More detailed information may be found in References 1, 2, and 3.

The elements iron, carbon, and chromium form the basic composition of stainless steels but significant amounts of many other elements are added, particularly to the precipitation-hardenable types, to provide a wide range of mechanical properties and formability. Each element in a stainless steel performs two functions in defining structure, one at elevated temperature and the other on cooling from elevated temperature. The function of the elements at elevated temperature will be discussed first.

The two crystallographic arrangements of stainless steels at high temperatures are ferrite, a body-centered cubic structure, and austenite, a face-centered cubic structure. Each element in a stainless steel composition tends to promote the formation of one or the other of these structures. The ferrite promoting elements are chromium, molybdenum, vanadium, columbium, silicon, aluminum, titanium, and phosphorus. The austenite promoting elements are iron, carbon, nickel, manganese, nitrogen, copper, and cobalt. Thus, the structure of a stainless steel at elevated temperature depends on the relative proportion of ferrite and austenite promoting elements in the composition. Steels that are ferritic at high temperature retain this structure on cooling and are not heat treatable. The steels that are austenitic at high temperature may remain austenitic on cooling or may transform to martensite, a body-centered tetragonal structure, with higher strength.

The symbol, M_s, is used to indicate the temperature at which the transformation of austenite to martensite begins and the symbol, M_f, is used to designate the temperature at which transformation ends. All of the elements used in stainless steels, with the exception of aluminum and cobalt, tend to lower the M_s and M_f temperatures of compositions that are austenitic at high temperatures. Therefore, stainless steels with high alloy content remain austenitic on cooling, while those with comparatively low alloy content will transform to martensite. Both of these structures can be strengthened by precipitation hardening.

The solubility of the alloying elements in austenite increases with increasing temperature. Therefore, in stainless steels of intermediate alloy content, the M_s and M_f temperatures can be controlled. At high annealing temperatures, the alloy content of the austenite is increased and, on cooling, the M_s temperature is depressed. At lower annealing temperatures the austenite is leaner in alloy content and, on cooling, transforms to martensite. Stainless steels that behave in this manner are called semiaustenitic; they are also precipitation hardenable. Figure 1 illustrates the manner in which alloy content affects the transformation temperature of stainless steels that are austenite at their annealing temperature.

Precipitation hardening is defined as the strengthening of a material by the precipitation of a constituent from a supersaturated solid solution. Small amounts of aluminum, titanium, copper, or phosphorus added to stainless steels are believed to be the elements

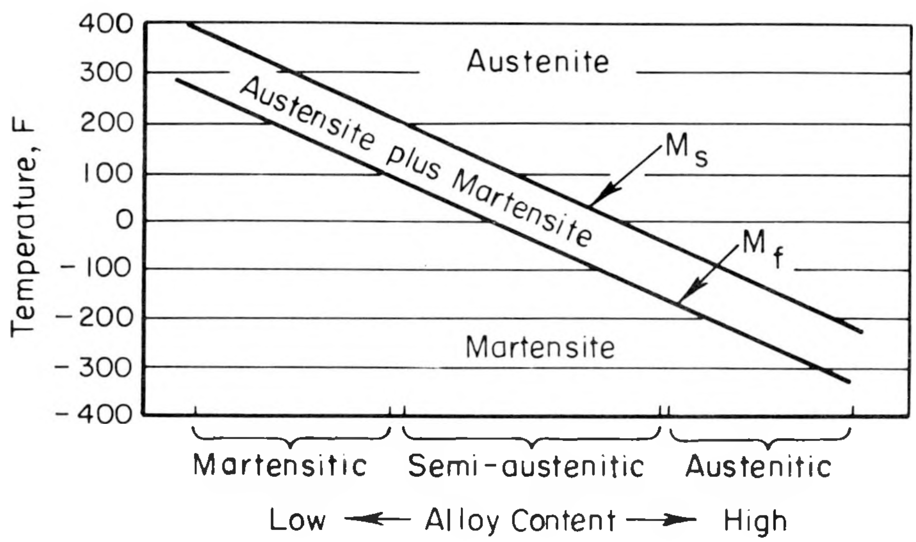

FIGURE 1. THE EFFECT OF ALLOY CONTENT ON THE TRANSFORMATION TEMPERATURE OF PH STAINLESS STEELS

responsible for precipitation hardening. There are three steps in the complete heat-treating cycle of precipitation-hardenable stainless steels.

The first step is solution heat treatment. In this step the stainless steel is soaked at elevated temperature to dissolve significant amounts of the hardening elements in the austenite matrix. The exact time and temperature of the treatment depend on the composition of the alloy and the mechanical properties desired. In addition, this step also is used to adjust the austenite composition of semiaustenitic and martensitic stainless steels and, thus, control the transformation temperature.

The second step is the formation of a supersaturated condition. Because most elements are less soluble at room temperature than at elevated temperature, supersaturation of the hardening elements is achieved by rapid cooling or quenching. An even greater condition of supersaturation can be obtained if the alloy transforms to martensite because the alloying elements are less soluble in martensite than in austenite. This transformation can be accomplished by subzero cooling to the M_f temperature or through control of the austenite composition by the solution treatment so that transformation is complete at room temperature.

The third step is precipitation hardening or aging. Precipitation does not occur naturally at room temperature in stainless steels because diffusion rates are negligible. However, atom migration takes place at somewhat higher temperatures and intermetallic compounds are formed from the alloying elements. These submicroscopic precipitates occur along the crystallographic planes of the matrix material. However, differences in lattice dimensions between the precipitate compound and the matrix cause the matrix to become severely strained. The development of this strained condition accounts for the strengthening that is brought about by the precipitation reaction.

By varying the aging time and temperature, the size and distribution of the precipitates can be controlled. Long exposure at temperatures low in the precipitation range is most effective for strengthening because the precipitate particles are small and uniformly distributed. At intermediate temperatures, the precipitate particles are somewhat larger and are not quite as effective but maximum strengthening can be achieved in a much shorter time. At temperatures high in the precipitation range the particles grow to even larger size, shearing takes place between the precipitate and matrix relieving the strain, and the material is considered to be overaged.

THERMAL TREATMENTS

The thermal treatments used to process the precipitation-hardenable stainless steels include annealing, homogenizing, solution treating, austenite conditioning, refrigerating, and precipitation hardening. Because of differences in composition, crystallographic structure and response to heat treatment, the processing of the three types of alloys is discussed separately in the following sections.

MARTENSITIC ALLOYS

The martensitic precipitation-hardening stainless steels develop their properties by two mechanisms. The first is by transformation of austenite to martensite during cooling from the annealing temperature, and the second by the precipitation of the hardening elements as intermetallic compounds on reheating to lower temperatures.

These alloys, at their normal annealing temperatures, have essentially an austenitic structure although some, because of their composition balance, may contain small amounts of the high-temperature delta ferrite phase. The presence of delta ferrite is undesirable because it provides a two-phase structure which is inherently low in ductility and toughness, particularly in transverse directions.

The factors most critical in the thermal processing of the martensitic precipitation-hardenable stainless steels are: the annealing temperature, the cooling conditions and the precipitation-hardening temperature. How these factors affect metallurgical structure and mechanical properties is discussed in the following sections.

Annealing and Solution Treatment. The annealing temperatures of the individual martensitic precipitation-hardenable stainless steels are selected to give the optimum combination of austenite composition for subsequent transformation to martensite and solution of hardening elements for precipitation in the martensite when reheated to the aging temperature. Annealing at temperatures higher or lower than the recommended temperatures will affect both the metallurgical structure and properties.

When temperatures lower than normal are used to anneal the martensitic types of precipitation-hardenable stainless steels, both the yield strength and ultimate strength will be reduced when the material is aged. The reasons for the lower strengths are that lesser amounts of the hardening elements are taken into solution and that a softer martensite is formed because it is lower in carbon content. Annealing at temperatures higher than normal will result in high ultimate strength but low yield strength. At higher temperatures, greater amounts of alloying elements are taken into solution and less transformation takes place on cooling. The retained austenite accounts for the lower yield strength. However, when the yield load is exceeded by deformation, this austenite transforms to stronger martensite and the ultimate strength is higher.

The ductility of the martensitic precipitation-hardenable stainless steels also can be adversely affected by high annealing temperatures. The compositions of most of these alloys are chemically balanced so that they are completely austenitic at normal annealing temperatures. When these temperatures are exceeded, delta ferrite formation may take place. This phase is retained on cooling to room temperature and, on aging, carbides are precipitated at the delta ferrite boundaries causing low ductility and poor impact strengths.

Castings and billets and forgings of large cross section are sometimes given an homogenizing treatment. This consists of soaking the alloys at temperatures 200 to 300 degrees higher than the normal annealing temperature so that the alloying elements are more uniformly distributed by diffusion. After cooling from the homogenizing temperature, the alloys must be annealed in the normal manner so that they will respond properly to the precipitation-hardening treatment.

Care must be exercised in heating martensitic precipitation-hardening alloys of large cross section to either the annealing or homogenizing temperatures. Charging thick sections into a hot furnace may cause internal cracking. Large parts should be heated uniformly at 1200 to 1400 F before the temperature is raised for annealing or homogenizing.

Cooling and Transformation. Martensitic precipitation-hardenable stainless steels are continuously cooled from their annealing temperature to room temperature. As they are cooled, the elements taken into solution at the annealing temperature become less soluble and a supersaturated solid solution results. In addition these alloys, because of their composition, transform above room temperature from austenite to martensite, increasing the degree of supersaturation.

A volume increase occurs with the transformation of austenite to martensite. Therefore, cooling from the elevated temperatures used for annealing, homogenizing or forging should be controlled to prevent cracking from the mechanical strains caused by this expansion. Small parts may be water or oil quenched safely, but larger sections should be air cooled, and cooling under cover is required for extremely massive parts. Also, it is essential that the cooling be continued to below the M_f temperature to assure complete transformation to martensite so that proper response to the subsequent age-hardening treatment is obtained.

Precipitation Hardening. The structure of annealed martensitic precipitation-hardenable stainless steel is an untempered martensite supersaturated with certain hardening elements. The final step in heat treating alloys of this type consists of reheating to intermediate temperatures in the range of 850 to 1150 F. This operation improves ductility by tempering the martensite and raises both yield and ultimate strengths by precipitation of the hardening elements as intermetallic compounds in the martensite.

The size and distribution of the precipitate particles can be controlled by varying the temperature and time of the precipitation treatment. By this means a wide range of mechanical properties may be developed. Highest strengths are obtained when small, uniformly distributed particles are formed by holding for long times at temperatures low in the precipitation range. Aging at intermediate temperatures forms larger but fewer precipitate particles and results in lower strengths and greater ductility. When temperatures high in the precipitation range are used, the precipitate particles become so large that they are ineffective in strengthening. This is called an overaged condition.

The overaged condition of the martensitic precipitation-hardenable stainless steels is recommended for cold heading and for room-temperature sheet-forming operations. The tensile yield and ultimate strengths will be about the same as for the annealed condition but overaging greatly improves ductility. However, if high strengths are required, the finished parts must be heat treated again through both the annealing and aging cycles.

SEMIAUSTENITIC ALLOYS

The thermal treatments for the semiaustenitic alloys include annealing, austenite conditioning, refrigeration, and precipitation hardening. Annealing is used to provide the most formable condition and to relieve residual stresses from cold working. Austenite conditioning and refrigeration accomplish the austenite-to-martensite transformation. Precipitation hardening produces the final desired mechanical properties.

Annealing and Solution Treatment. The semiaustenitic precipitation-hardenable stainless steels are annealed by heating at high temperature followed by rapid cooling to room temperature. The major purposes of annealing are to put the steel in its most formable condition and to take into solution the alloying elements responsible for hardening in subsequent precipitation treatments. Also, annealing treatments are often used to restore formability when severe forming operations, such as spinning and deep drawing, have caused transformation to martensite by cold work.

Variations from the normal annealing temperature can affect the formability of the semiaustenitic steels. When temperatures below normal are used, a less stable austenite is formed which will have higher tensile properties and reduced ductility. Higher than normal annealing temperatures can cause grain growth and increased amounts of delta ferrite which tend to lower elongation values.

Annealing at temperature above or below the recommended temperatures also can adversely affect properties after subsequent heat treatment. When annealing temperatures are too high, lesser amounts of martensite are formed and low strength in the fully hardened condition will result. Low annealing temperatures will cause low elongation values after transforming and aging the semiaustenitic precipitation-hardening alloys.

Austenite Conditioning and Transformation. The temperature at which the austenite-to-martensite transformation takes place depends on the composition of the austenite when cooling begins. The decreasing solubility of chromium and carbon in austenite, as temperature is lowered, permits the adjustment of the M_s-M_f temperature range of the semiaustenitic alloys. The two standard austenite conditioning treatments used for alloys of this type are: a low-temperature treatment where a lean austenite results from formation of large amounts of chromium carbide; and, a higher temperature conditioning treatment which takes more chromium and carbon into solution and so requires refrigeration to complete transformation to martensite on cooling.

The low-temperature austenite conditioning treatments were the first used for heat treatment of semiaustenitic precipitation-hardenable steels. At these conditioning temperatures significant amounts of chromium and carbon are rejected from the austenite and form chromium carbides at the sites of highest internal energy, usually austenite grain boundaries and austenite-delta ferrite phase interfaces. The depletion of chromium near the grain boundaries promotes susceptibility to intergranular attack in corrosive environments. Thus, when a steel in this condition is to be pickled in an acid bath, the time in the solution must be carefully controlled; if too long, the grain boundaries will have been preferentially attacked.

Higher strengths are attainable when the high-temperature austenite conditioning treatments are used for the semiaustenitic alloys. Higher carbon content of the martensite formed probably accounts for the strength difference. However, higher alloy content depresses the transformation temperature and refrigeration at -100 F is required to obtain a maximum amount of martensite. Low strengths can result because of austenite retention if processing is not carefully controlled. If the conditioning temperature is too high or if the material is held between the M_s and M_f temperature on cooling, stable austenite may form which will not transform to martensite even when refrigerated at -100 F.

Austenite conditioning temperatures intermediate to the standard heat treatments are often used in the fabrication of brazed assemblies. The temperature employed under these conditions is selected to be compatible with the flow temperature of the brazing alloy. Massive brazing tools and fixtures may cause slow cooling rates which can affect both transformation and subsequent precipitation. Adjustment of refrigeration and aging temperatures to obtain desired mechanical properties may be necessary in such instances.

Precipitation Hardening. Precipitation hardening of the semiaustenitic stainless steels is accomplished by reheating at temperatures of 850 to 1050 F. In this step, the martensite is tempered, improving ductility, and intermetallic compounds are precipitated providing further strengthening. Each alloy of this type reaches a peak strength at a certain temperature but may be overaged at higher temperatures for improved toughness with some sacrifice of strength.

AUSTENITIC ALLOYS

The austenitic stainless steels are annealed and precipitation hardened. They are highly alloyed and do not transform when cooled from high temperatures. Also, because no transformation strains are present in the annealed structure, higher aging temperatures are required for effective strengthening by precipitation.

<u>Annealing and Solution Treatment</u>. The austenitic precipitation-hardenable stainless steels are annealed by heating at high temperatures to dissolve the hardening elements in the austenitic matrix followed by rapid cooling to obtain a supersaturated solid solution. The annealing temperatures are selected according to the mechanical properties desired. Low annealing temperatures provide the highest room-temperature strength and ductility, while higher annealing temperatures result in better elevated-temperature properties but ductility may be reduced because of increased grain size.

<u>Precipitation Hardening</u>. The austenitic precipitation-hardenable stainless steels are strengthened by heating at temperatures around 1300 F for times as long as 16 hours. The high temperatures and extended times are necessary to reach peak hardness because of the lower degree of supersaturation in the austenitic matrix and because there is no internal strain from martensite transformation to force the precipitation reaction.

Deformation from room-temperature forming operations will increase the internal energy available for precipitation, and maximum strengths can be reached in cold-worked material by aging at lower temperatures. Fabricated parts that are severely cold worked in localized areas during forming may be reannealed before precipitation hardening to prevent nonuniform response to heat treatment. If the deformation is uniform as in some stretch forming operations, the parts may be hardened by aging directly at selected lower temperatures.

MECHANICAL TREATMENTS

The precipitation-hardenable stainless steels are shaped into usable parts by both hot working and cold working. These processes can further be described as primary and secondary forming operations. Primary operations include forging, rolling, extruding, and tube and wire drawing, while typical secondary operations are spinning, bending, deep drawing, stretching and drop hammer forming. A previous NASA Technical Memorandum (Ref. 4), describes in detail the equipment and procedures used in the deformation processing of precipitation-hardenable stainless steels. The purpose of this section of the present report is to show how variations in temperature and in the degree of deformation during such processing will affect the metallurgical structures and properties of the precipitation-hardenable stainless steels.

HOT WORKING

Although the precipitation-hardenable stainless steels are used largely as wrought products, all are at some time in cast form. The primary deformation processes such as forging or rolling, in addition to producing a desired shape, are essential in producing proper metallurgical structures. The hot-working temperatures and percent reductions greatly affect the final mechanical properties of the precipitation-hardenable stainless steels by their influence on chemical homogeneity, metallurgical structure, and grain size.

Process factors that must be controlled in hot working to produce satisfactory material include: heating rate to the working temperature, uniform through heating, temperature at the beginning of hot working, temperature control during working, finishing temperature, the amount of final reduction, and cooling rate from the finishing temperature.

Large cross sections must be heated slowly to the hot-working temperatures. High internal stresses may cause cracking when thick pieces are charged into a very hot furnace. The piece should be held at temperature until the full cross section is uniformly heated before hot working is started. This reduces the possibility of cracking during working because the mechanical properties are the same throughout the part. If the center of a piece is too cold, it may not receive enough work to break up the cast dendritic structure. This can result in poor response to subsequent heat treatment and low ductility in finished parts.

The upper limit of the hot-working temperature range should not be exceeded. If parts are overheated before hot working, or if the energy of the working causes the part temperature to rise, delta ferrite will form and may cause cracking during further hot working. In addition, the delta ferrite will be retained on cooling to room temperature and will remain after subsequent heat treatment with an adverse effect on transverse ductility.

The finishing hot-working temperature must be high enough that the materials are not cracked by hot working but must be closely controlled so that, with the proper amount of final reduction, satisfactory grain size is achieved. Control of cooling rate after hot working to prevent cracking is necessary especially for the martensitic types of precipitation-hardenable alloys that exhibit a volume change during transformation.

COLD WORKING

Wrought precipitation-hardenable stainless steels are cold worked by many deformation processes. Primary methods, like cold rolling and cold drawing are used to enhance mechanical properties. These alloys also are formed into usable shapes by many secondary deformation processes. The success of these operations and the quality of the finished product depend on both the condition of the material when it is cold worked and the details of subsequent thermal treatments. The interaction of these factors is discussed separately for martensitic, semiaustenitic, and austenitic types of precipitation-hardenable stainless steels in the following sections.

Martensitic Alloys. The precipitation-hardenable steels of the martensitic type may be cold worked at room temperature in either the annealed or overaged conditions. The condition in which the material is worked is governed by the properties of the martensite. The alloys low in carbon content form a relatively ductile untempered martensite that may be worked severely. Other alloys of greater carbon content have higher strength with lower ductility and must be overaged for successful room-temperature cold working.

Martensitic alloys cold worked in the annealed condition and subsequently precipitation hardened will have tensile properties higher than those obtained by aging annealed material. The extent of the strength increase depends on the amount of deformation and the cold-working temperature. If a part is not uniformly worked during fabrication, it may exhibit considerable variation in strength and hardness between heavily worked and lightly worked areas. Completed parts may be reheat treated by annealing and aging if uniform mechanical properties are desired.

Parts that are fabricated from the martensitic precipitation-hardenable stainless steels in the overaged condition must be reannealed prior to aging when higher strengths are required. If the strength in the overaged condition is acceptable, parts that are severely formed should be reaged at the proper temperature to restore toughness and to relieve residual stresses. Complete reheat treatment is recommended for parts that must be free from all residual stress.

It is difficult to deep draw some martensitic alloys in either the annealed or overaged condition. However, forming by this method may be accomplished during cooling from the annealing temperature while the steel is still austenitic. The forming operation should be completed before the temperature of the part reaches about 300 degrees above the normal M_s temperature of the alloy. If forming is continued at lower temperatures, failures may occur because the deformation will induce the transformation of austenite to martensite which has much lower ductility. The warm-formed parts must be cooled to room temperature to assure complete transformation to martensite. Subsequent precipitation hardening will give mechanical properties comparable to those obtained by standard heat treatment.

Semiaustenitic Alloys. The semiaustenitic precipitation-hardenable stainless steels are normally cold worked in the annealed condition. The major effect of the mechanical deformation is transformation of the austenitic matrix to martensite. Wrought products are cold reduced up to 60 percent by primary deformation processes. After this uniform heavy cold working, the hard martensite need only be aged at low temperature to obtain very high strength.

Maximum strength is developed in material transformed by severe cold reduction on aging at lower temperatures than those used when the martensitic structure is formed by heat treatment. The reason is that the cold work introduces internal strain energy and, thus, less thermal energy is required to drive the precipitation reaction.

The major portion of the semiaustenitic stainless steel produced is fabricated in the annealed condition by secondary deformation processes. Because of the low strength and high ductility they display in this condition, these alloys can be shaped into complex parts by methods such as deep drawing, spinning and stretch wrapping. After fabrication the parts are heat treated to high strength by austenite conditioning at either high or low temperature followed by aging.

Parts may be nonuniformly cold worked during fabrication. Localized areas that have been work hardened may respond to subsequent heat treatment in a different manner than unworked areas. The higher temperature austenite conditioning treatments are more effective in relieving residual fabrication stresses and will produce more uniform mechanical properties throughout the part. Reannealing before heat treatment is recommended when complete uniformity of properties is required in parts that have been severely cold worked during fabrication. Reannealing, i.e., intermediate annealing, also is recommended to restore ductility when forming is severe and fabrication must be done in several steps.

Limited forming and sizing operations can be done on cooling from austenite conditioning temperatures. Formed parts can be held to close dimensional tolerance if they are restruck in the forming dies on cooling from the austenite conditioning temperature and held until the martensite transformation is complete. Heavy cold working after austenite conditioning will result in higher strength and low ductility. Reaustenite conditioning will restore the ductility of fully heat-treated parts.

Some semiaustenitic alloys show a loss in strength when they are cold worked in the annealed condition and subsequently heat treated by the low temperature austenite conditioning processes. Modified heat treatments developed to overcome the drop in properties after cold working employ slightly higher austenite conditioning temperatures and refrigeration before precipitation hardening.

Austenitic Alloys. Austenitic precipitation-hardenable stainless steels can be satisfactorily cold worked in the annealed condition by both primary and secondary deformation processes. The austenitic structure is not transformed by cold work but the mechanical deformation will influence the response to aging. Lower precipitation-hardening temperatures may be used to achieve maximum strength after cold working because the additional internal strain energy will accelerate the aging response.

Nonuniform properties in shaped parts having deformed and undeformed areas can be avoided by reannealing after fabrication and prior to aging. Another method to minimize nonuniform heat treat response is to overage at slightly higher than normal temperature for a period of time to equalize strain energy, then lower the temperature and precipitation harden in the usual manner.

The austenitic age hardenable alloys are susceptible to grain growth if they are cold worked small amounts, about 1 to 4 percent, and then annealed. Large grain size may have a harmful effect on creep and fatigue properties. Therefore, nonuniformly cold-worked parts that may have areas deformed to these critical amounts should not be reannealed after forming.

HEAT-TREATING EQUIPMENT

The types of furnaces used to heat treat precipitation-hardenable stainless steels depend on the product to be processed. For example, continuous furnaces may be used for annealing of strip. The details of their construction and operation are not considered here, but the characteristics necessary for proper thermal processing of the precipitation-hardening alloys are discussed. Complete information on heat-treating equipment and procedures can be found in Ref. 5.

As described previously, the purpose of each step in heat treatment is to produce a specific condition in the material. The final objective of the combined processing is a high-strength corrosion-resistant part. To achieve this goal, primary consideration must be given to furnace atmosphere and control of temperature.

Air has proven to be the most satisfactory atmosphere for heat treating precipitation-hardenable stainless steels. Electric furnaces or gas-fired radiant tube furnaces are generally used. Open gas or oil-fired furnaces present the problems of contamination from combustion products and flame impingement on the work. Controlled reducing atmospheres are not recommended because they may carburize or nitride surfaces of the parts being heated. Both carbon and nitrogen are strong austenite promoters. Contamination by these elements can adversely affect the response to heat treatment of the delicately balanced precipitation-hardenable stainless steel compositions. Such contamination also will lower the corrosion resistance of these alloys.

Bright annealing may be done in dry hydrogen, argon, or helium with a dew point of approximately -65 F. These atmospheres may be used also for the higher temperature, 1700-1750 F, austenite conditioning treatments of the semiaustenitic alloys if the dew point of the gas is -75 F or lower. Vacuum furnaces are necessary for scale-free heat treatment at lower temperatures. When dry gas or vacuum atmospheres are used, the cooling rate from the heat-treating temperature to 1000 F should be approximately the same as an air cool to put the material in the desired condition.

Molten salt baths rectified with carbonaceous material to make them neutral to low alloy or tool steels should not be used to heat treat the precipitation-hardenable alloys. The molten salts can cause intergranular attack and the alloys will be carburized with harmful effects on mechanical properties and corrosion resistance.

Accurate control of temperature is essential in all thermal processing of the precipitation-hardenable stainless steels. Common temperature tolerances are $\pm$ 25 F for annealing, $\pm$ 15 for austenite conditioning and $\pm$ 10 F for refrigeration and precipitation hardening. Furnace thermocouples should be calibrated before installation, checked daily and replaced regularly. In addition, the uniformity of the temperature in the furnace should be checked on a routine schedule so that burned-out heating elements or defective burners may be discovered and replaced.

Commercial refrigeration equipment can be obtained to perform the -100 F transformation step in the treatment of semiaustenitic alloys. However, a temperature of -100 to -109 F can be maintained in a bath of alcohol or acetone containing an excess of dry ice (solid carbon dioxide).

CLEANING, DESCALING, AND PICKLING

The precipitation-hardenable stainless steels should be thoroughly cleaned before heat treatment to remove all traces of machining coolants or forming lubricants. The removal of the surface grease or oil prevents carburization and facilitates scale removal after heat treatment. Water soluble lubricants can be removed with alkaline cleaners. Vapor degreasers or solvent cleaners should be used for mineral lubricants. The work pieces should be washed in hot water and dried after cleaning and before heat treatment.

When the precipitation-hardenable stainless steels are formed over Kirksite dies, zinc alloy particles may become embedded in the part surfaces. This material must be removed before heat treatment because at elevated temperatures the zinc will penetrate along grain boundaries and cause embrittlement. Embedded or adhering die particles may be removed by cleaning the formed parts in a nitric acid solution before heat treatment.

Scale will develop on the precipitation-hardenable stainless steels during heat treatment in air. The amount of scale formed depends on the heat-treating temperature and the time of exposure. Both abrasive and chemical methods can be used to descale these alloys. Detailed descriptions of cleaning and descaling procedures and equipment are given in References 6 and 7.

Both wet and dry abrasive blasting methods are used successfully for descaling. Common grit sizes are No. 36 for dry blasting and No. 220 for wet blasting. Grit materials used are silica or alumina. The air pressures employed vary from 25 to 100 psi with the lower pressures used on thin gage material. Abrasive methods can be used without harmful effects on the metallurgical condition of the material. If the same abrasive cleaning equipment is used to clean carbon and alloy steels, the grit should be changed before blasting stainless steel parts to avoid contaminating the surfaces with iron particles. All work should be pickled after grit blasting in a 10 percent nitric - 2 percent hydrofluoric acid solution to remove any residual scale or embedded contaminants. Pickling in this solution must be closely controlled to prevent excessive metal removal, etching, and intergranular corrosive attack. The bath temperature should not exceed 140 F and pickling time should be held to a maximum of three minutes. After pickling, the steel should be water rinsed and dried.

The susceptibility of these steels to intergranular corrosion during pickling depends on alloy composition and prior heat treatment. The semiaustenitic alloys in either the annealed or cold-rolled and aged conditions, and the austenitic alloys, are the most resistant to intergranular attack. However, the semiaustenitic alloys in other conditions of heat treatment, and the martensitic alloys, are subject to severe intergranular penetration in acid environments. Depletion of chromium at the grain boundaries lowers the corrosion resistance of these areas. This occurs when chromium carbides are formed during thermal processing to obtain high strengths. Consequently, the immersion time in the pickling solutions should be as short as possible when these alloys are in intermediate or fully heat-treated conditions.

Conditioning treatments are sometimes necessary before pickling to soften the tightly adhering scale formed during high-temperature heat treatments. Scale-softening methods that are used include immersion in caustic permanganate solution, sodium hydride baths, and several proprietary fused salt baths. The caustic permanganate solution, 18 to 20 percent NaOH - 4 to 6 percent $KMnO_4$, is used at about 190 F while the operating temperatures for the other methods range from 700 to 1100 F. Because of its relatively low temperature, the caustic-permanganate method may be used on the precipitation-hardenable steels in any condition of heat treatment. However, the higher temperatures employed by the other methods can affect the metallurgical structures, heat treat response and properties of these alloys. For example, if semiaustenitic alloys are processed at these temperatures after austenite conditioning at 1700-1750 F, full strength will not be developed after refrigeration and aging because of austenite stabilization. Low strength also may result from overaging if fully heat-treated alloys are scale conditioned at temperatures higher than the normal precipitation-hardening temperature. Combined scale-conditioning and precipitation-hardening treatments may be possible. However, the operating characteristics of the high-temperature scale-conditioning baths, particularly temperature uniformity and control, should be established before processing in this manner.

EFFECT OF THERMAL AND MECHANICAL TREATMENTS ON THE MECHANICAL PROPERTIES OF THE ALLOYS

The basic and determining influence of thermal and mechanical treatments on the mechanical properties of the PH stainless steels has been emphasized in the preceding sections of this memorandum. It has been noted that the operations that may be involved in arriving at the desired properties for each alloy include hot or cold working, annealing or solution treating, austenite conditioning, transformation of austenite to martensite, precipitation hardening and tempering. Appropriate cooling steps follow the heating operations. These operations all affect the structure of the alloys in one way or another, and thereby determine the mechanical properties that are obtained. The specific effects of each thermal or mechanical treatment, and the mechanisms by which they influence the structure and properties of the alloys have been described in the section on "Physical Metallurgy".

Not all of the above operations are necessarily used to arrive at desirable conditions for each alloy. The operations that are necessary, and their details, depend on the type of PH steel (i.e., martensitic etc.) and the mechanical properties that are desired. Standard treatments usually involve several of the steps mentioned above and it is important to know the effect of each step on the properties so that the alloy may be properly processed and used. The properties that are developed following intermediate stages in a treatment will govern the nature of fabricating or processing operations that may follow. Therefore tabulations of mechanical properties usually include data for the alloys in various intermediate "conditions". Such conditions are identified by combinations of letters, numbers or words (depending on the producer of the alloy) that indicate the nature of the treatment that has been given to the alloy up to that point in the process.

In the present section of the report, mechanical properties are given for each alloy in a number of standard conditions illustrating the effect of the thermal and mechanical processing steps that are used to arrive at the respective conditions. During the development and practical use of the alloys, numerous investigations have been conducted in which the details of the treatments have been varied, either in attempts to improve the alloys, or to develop specific mechanical properties differing from those obtained by standard treatments. Also, a considerable amount of work has been done to determine the effects of deviations from standard heat treatment details on mechanical properties, in order to establish practical operating limits and procedures. The data from these investigations have been assembled for each alloy in subsequent pages so that related information on the effect of the thermal and/or mechanical treatments on the mechanical properties of the alloys may be readily available for comparison.

MARTENSITIC PH STAINLESS STEELS

17-4 PH. This alloy may be obtained in bar, plate, sheet, wire, billet and cast form and is given the specific treatments at the mill designed to place the material in a suitable condition for subsequent fabricating operations. Most forms are usually supplied in the solution-treated Condition A. This is an anneal at 1900 ± 25 F for 30 minutes followed by air cooling. Austenite is transformed into martensite on cooling to room temperature. The martensite formed on cooling is supersaturated with copper, and this element precipitating within the martensite during the aging treatments provides the strengthening.

Fabricating operations may be conducted on the alloy in the annealed condition, and aging treatments may follow directly after fabrication. A number of aging treatments have been developed to provide a wide range of properties. The treatments are listed in Table II, and the effects of variation in aging temperature on typical mechanical properties of cold-flattened sheet and strip are shown in Table III. The large difference in tensile and yield strengths between Conditions A and H 900 is the result of the precipitation-hardening mechanism. The strength and hardness decrease gradually with increasing temperature of aging until the overaged condition is approached at about 1150 F. In the H 1150 condition there is a definite increase in elongation. The H 1150 and H 1150-M treatments may be used intentionally to improve formability in certain operations such as stretch forming (Ref. 8). After such treatment, however, parts must be reannealed at 1900 F prior to aging to any of the higher strength conditions, when such strengths are required.

The treatments applied to bar stock are the same as for sheet. Table IV shows the effect of the various aging treatments on the mechanical properties developed in 1-inch bars. The strength closely matches that obtained on sheet and strip, but the bar material is somewhat more ductile. The change in room-temperature tensile properties with increasing aging temperature is also illustrated in Figure 2.

The elevated-temperature properties also vary with variation in the aging temperature as shown in Figure 3 where the H 900 and H 1150 short-time tensile properties are compared at test temperatures from 600 to

TABLE II. HEAT TREATMENTS FOR 17-4 PH (Ref. 8)

Condition A

Solution treated
1900 F 1/2 Hour
Air Cool

Condition	Heat to $\pm$ 15 F	Time at Temperature, hours	Type of Cooling
H 900	900 F	1	Air
H 925	925 F	4	Air
H 1025	1025 F	4	Air
H 1075	1075 F	4	Air
H 1150	1150 F	4	Air
H 1150-M	1400 F	2	Air
		Followed by	
	1150 F	4	Air

TABLE III. EFFECT OF HEAT TREATMENTS ON TYPICAL MECHANICAL PROPERTIES OF 17-4 PH SHEETS AND STRIP-COLD FLATTENED (Ref. 8)

Property	Condition						
	A	H 900	H 925	H 1025	H 1075	H 1150	H 1150-M
Ultimate Tensile Strength, ksi	160	210	200	185	175	160	150
0.2% Yield Strength, ksi	145	200	195	170	165	150	130
Elongation, % in 2"	5.0	7.0	8.0	8.0	8.0	11.0	12.0
Hardness-Rockwell	C 35	C 45	C 43	C 38	C 37	C 35	C 33

TABLE IV. EFFECT OF HEAT TREATMENT ON THE TYPICAL MECHANICAL PROPERTIES OF 1-INCH 17-4 PH BARS (Ref. 9)

	Condition					
	H 900	H 925	H 1025	H 1075	H 1100	H 1150
Ultimate Tensile Strength, psi	200,000	190,000	170,000	165,000	150,000	145,000
0.2% Yield Strength, psi	185,000	175,000	165,000	150,000	135,000	125,000
Elongation, % in 2" or 4 x D	14.0	14.0	15.0	16.0	17.0	19.0
Reduction of Area, %	50.0	54.0	56.0	58.0	58.0	60.0
Rockwell-Hardness	C44	C42	C38	C36	C34	C33
Impact, Charpy V-Notch, ft-lbs	20	25	35	40	45	50

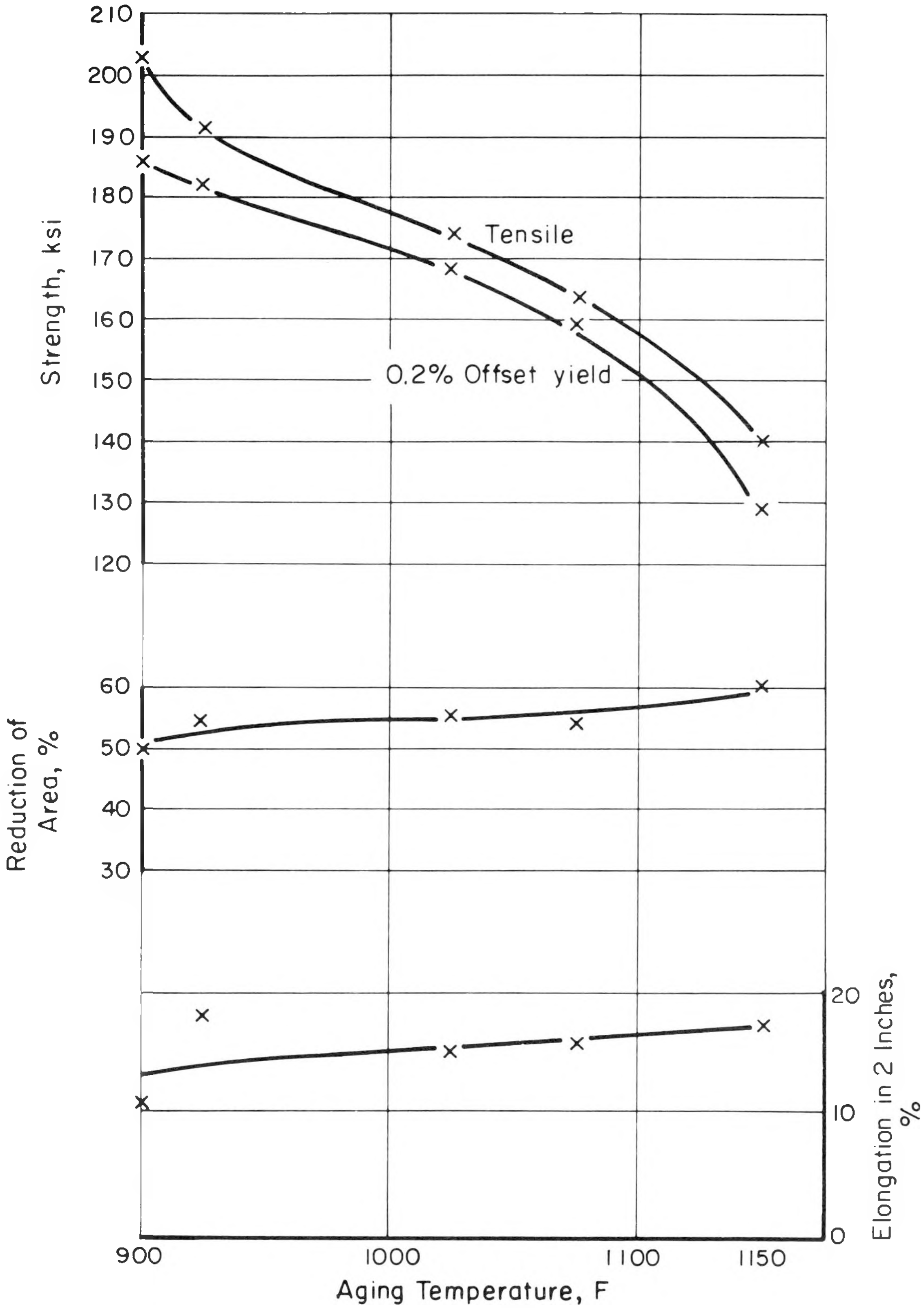

FIGURE 2. EFFECT OF AGING ON ROOM-TEMPERATURE PROPERTIES OF 1-INCH-DIAMETER 17-4 PH BAR (Ref. 9)

Solution treated 1/2 hour at 1900 F
Aged 1 hour at 900 F; 4 hours at other temperatures

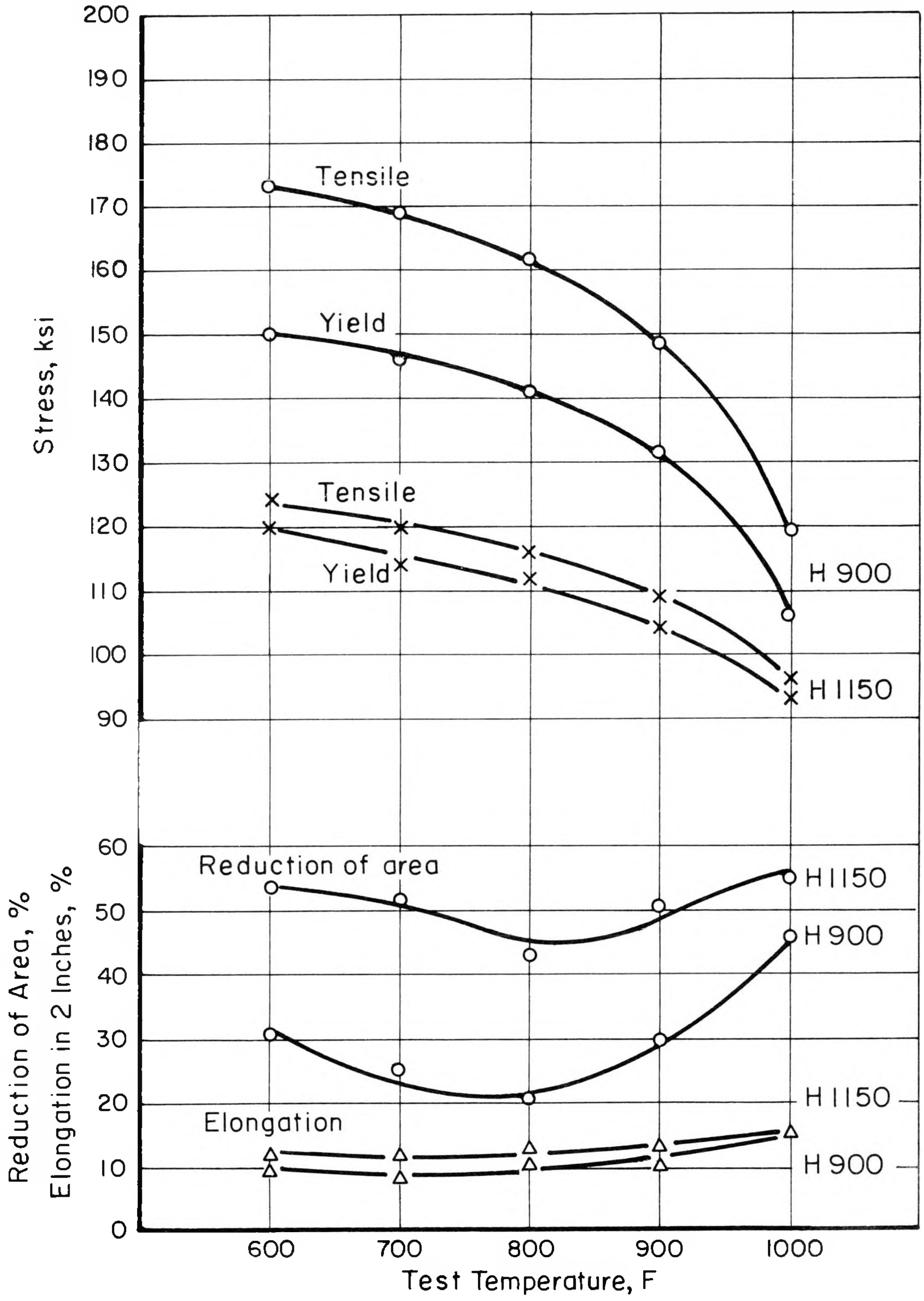

FIGURE 3. EFFECT OF AGING ON ELEVATED TEMPERATURE TENSILE PROPERTIES OF 1-INCH-DIAMETER 17-4 PH BAR (Ref. 9)

1000 F. The tensile and yield strengths decrease gradually for both conditions, but the ductility as measured by reduction of area appears to be at a minimum at about 800 F. The ductility in the H 900 condition shows a fairly large increase when tested at temperatures above 800 F. It approaches that of the H 1150 condition at the 1000 F test temperature.

The stress-to-rupture in 100 hours and 1000 hours, as influenced by the aging temperature, has been determined for 17-4 PH bar stock, and typical data are listed in Table V. In the fully hardened condition, H 900, the material is most resistant to rupture under stress at elevated temperatures.

The ductility of fully hardened 17-4 PH in large sections may be appreciably lower in a direction transverse to the direction of rolling, or major axis of forging, than in the longitudinal direction. This is particularly true in the short transverse direction, i.e., through the thickness. It has been found that aging at higher temperatures greatly improves the transverse ductility. Data showing the effect of the H 1025, H 1075, and H 1150 treatments in comparison with H 925 are given in Table VI (Ref. 10) and Table VII (Ref. 9). The comparative values of reduction of area indicate that significant improvement in the transverse ductility has been achieved by the overaging heat treatments. The ductility in the longitudinal direction is only slightly increased, and still remains considerably greater than in the transverse direction. It should be noted that the high-temperature aging treatments caused a corresponding decrease in the tensile and yield strengths.

Continued study of this problem has resulted in the development of a procedure that provides a similar improvement in transverse ductility without sacrificing the strength obtainable by the H 925 treatment.

TABLE V. EFFECT OF HEAT TREATMENT ON THE STRESS-RUPTURE PROPERTIES OF 1-INCH DIAMETER 17-4 PH BAR (Ref. 9)

	625 F	700 F	800 F	900 F
Stress to Rupture				
in 100 hrs., Stress ksi				
Condition H 900	--	156	140	95
Condition H 925	163	154	128	--
Condition H 1075	137	126	108	--
Condition H 1150	123	114	100	80
in 1000 hrs., Stress ksi				
Condition H 900	--	150	128	60
Condition H 925	160	151	121	--
Condition H 1075	134	123	103	--
Condition H 1150	122	111	94	71

TABLE VI. EFFECT OF OVERAGING ON TRANSVERSE DUCTILITY OF 17-4 PH (Ref. 10)

Plate Thickness, inch	Direction[(a)]	Condition[(b)]	Ultimate Tensile Strength, ksi	Yield Strength, 0.2% Offset, ksi	Elongation in 4XD, %	Reduction of Area,%
4	L	H 925	183	177	16.0	55.0
	ST		184	-	5.5	3.2
4	L	H 1025	163	160	17.3	57.0
	ST		163	159	12.0	28.0
4	L	H 1075	155	147	17.5	61.0
	ST		154	148	10.0	21.0
4	L	H 1150	148	136	20.0	63.0
	ST		148	134	13.0	31.0

(a) L = longitudinal; T = transverse; ST = short transverse.

(b) H Conditions all 4 hours at temperature indicated; 1900 F for 1 hour air cooled, A condition precedes all H treatments.

TABLE VII. EFFECT OF HIGH-TEMPERATURE AGING ON LONGITUDINAL AND TRANSVERSE PROPERTIES OF 17-4 PH (Ref. 9)

Size of Stock	Direction of Test	Condition	Ultimate Tensile Strength, ksi	0.2% Yield Strength, ksi	Elongation in 2", %	Reduction of Area,%
4" Slab	L	H 1025	163	160	17.0	58.0
	*T	H 1025	162	159	13.0	32.0
	L	H 1075	155	144	18.0	61.0
	*T	H 1075	154	147	12.0	27.0
6" Square	L	H 1075	160	157	17.5	62.0
	T	H 1075	158	155	14.0	40.0
	L	H 1150	147	139	20.0	63.0
	T	H 1150	147	136	15.0	44.0

* Transverse tests on 4" slab made through thickness (short transverse direction).

This involves an homogenizing treatment of 1 to 2 hours at 2150 F followed by air cooling to at least 90 F, then solution treating and aging by the standard H 925 procedure. Examples of the effect of this treatment sequence on the transverse properties of the 17-4 PH, H 925 condition are given in Tables VIII and IX (Refs. 9,10). The figures show that the improvement in transverse ductility is of the same order of magnitude as that obtained by overaging. Clarke and Garvin (Ref. 10) pointed out however, that the homogenization treatment appears to be most effective in sections of intermediate size, and is not always effective in larger sections, about 10 by 10 inches and over. A new alloy, 15-5 PH, has been developed for use in applications involving such large sizes. This alloy is discussed on subsequent pages.

Additional data on this point from an investigation conducted to determine whether "as-received" plate in the solution-annealed condition should be homogenized and reannealed prior to fabrication of fuselage lugs. Tensile, yield and reduction of area values are compared in Figure 4 after 3 different aging treatments on homogenized and non-homogenized material. The authors concluded that while some improvement in reduction of area was obtained by homogenization, the improvement was not as great as that shown in Table VIII. The results are not directly comparable with those in Table VIII because aging temperatures were different. However, it may be noted that the reduction of area in the nonhomogenized condition used as the control value was considerably different in the two investigations discussed. For example, in Ref. 9, the H 925, nonhomogenized value for reduction of area was given as about 5.0%, while in Ref. 11, for the H 900 condition it was about 29.0%. The increase to about 35.0% in both cases, is considerably different on a percentage basis.

TABLE VIII. EFFECT OF HOMOGENIZING 17-4 PH ON THE TRANSVERSE TENSILE PROPERTIES IN THE H-925 CONDITION (Ref. 9)

Size of Stock	Direction of Test	Heat Treatment	Ultimate Tensile Strength, ksi	0.2% Yield Strength, ksi	Elongation in 2", %	Reduction of Area, %
8" Square	T	1900 F, H 925	188	171	5.0	5.4
	T	Homogenized, 1900 F, H 925	188	170	12.5	33.2
4" Slab	L	1900 F, H 925	194	-	17.0	53.5
	*T	1900 F, H 925	195	-	4.5	5.0
	*T	Homogenized, 1900 F, H 925	195	-	10.5	26.0

* Transverse tests on 4" slab made through thickness. Homogenizing treatment, 1-2 hours, 2150 F, cool to 90 F.

TABLE IX. EFFECT OF HOMOGENIZING HEAT TREATMENT ON TRANSVERSE DUCTILITY OF 17-4 PH (Ref. 10)

Section Size, inch	Direction(a)	Location	Condition(c)	Ultimate Tensile Strength, ksi	Elongation in 4 X D, %	Reduction of Area, %
4-inch diameter	T	C	H 925	187	2.8	7.3
			Ho + A + H 925	190	5.0	7.0
4-inch-thick plate	ST	I	H 925	184	5.6	3.1
			Ho + A + H 925	195	11.5	26.0
6-inch-thick plate	ST	I	H 925	191	1.0	14.3
			Ho + A + H 925	193	10.0	24.5

(a) L = longitudinal; T = transverse; ST = short transverse.

(b) C = center; I = intermediate.

(c) H conditions all 4 hours at temperature indicated; 1900 F for 1 hour air cooled, A condition precedes all H treatments. Ho 2150 F for 1-1/2 hours air cooled.

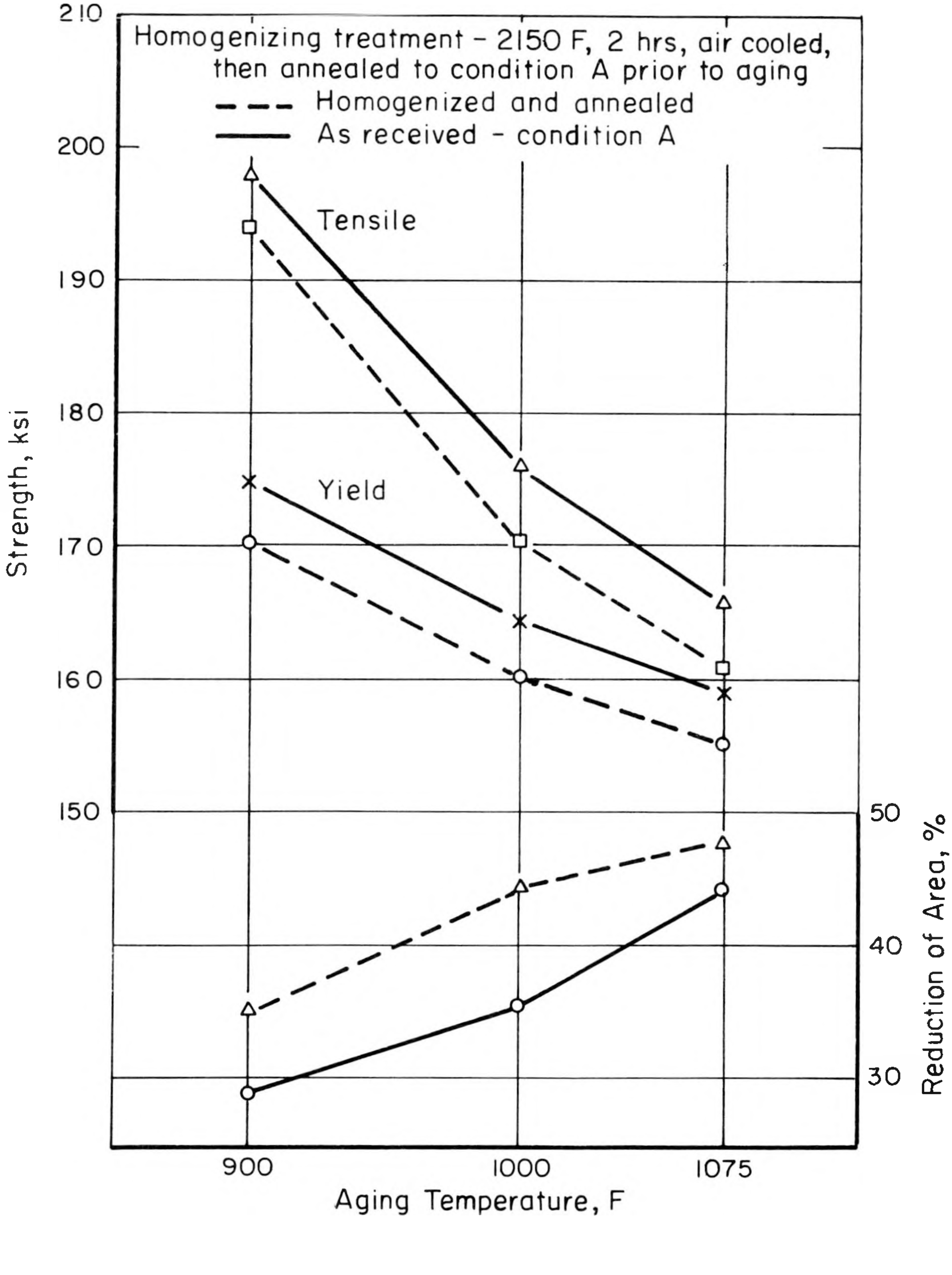

FIGURE 4. EFFECT OF HOMOGENIZING 17-4 PH PLATE PRIOR TO AGING (Ref. 11)

The influence of aging conditions on the impact properties of 17-4 PH bar from room to -320 F is shown in Table X. The higher the aging temperature the better the impact strength at all temperatures. At -110 F, the H 1150 condition still exhibits good impact strength, but at -320 F the strength is quite low for all three aging conditions.

A new heat treatment, designed to develop improved machinability and toughness is being evaluated, and shows promise of providing considerably better impact properties at -320 F, at some sacrifice in other mechanical properties. The new treatment is a double aging, first at 1300-1500 F for two hours and air cooling, followed by a 1150 F treatment for four hours and air cooling. The Charpy V-notch values following this treatment are 100-109 ft-lbs at room temperature and 29-27 at -320 F. This compares with values of 95 and 6.5 ft-lbs, respectively, for the H 1150 condition (Ref. 9).

Tensile tests conducted at intermediate temperatures during cool-down from the 1900 F solution-treatment temperature showed excellent ductility in the temperature range of 650-900 F (Ref. 8). Elongation in 2 inches reached a maximum value of 85 percent at 800 F. This favorable ductility may be utilized in hot forming in the 650-900 F range while cooling from the annealing temperature. At these hot-forming temperatures the steel is still austenitic, and does not transform to martensite until the Ms temperature is reached, i.e., below about 300 F. Mechanical properties of hot-formed parts subsequently age hardened are not significantly different from those obtained by standard heat treatments. This is shown by the data in Table XI. The properties listed are those obtained at room temperature on specimens that had been stretched the indicated amounts at 500, 650 and 800 F, cooled to room temperature, and tested in both aged and unaged conditions. The properties obtained by the standard H 900 treatment are given for comparison.

TABLE X. VARIATION IN IMPACT STRENGTH WITH CHANGES IN HEAT TREATMENT OF 17-4 PH BAR (Ref. 9)

Condition	Charpy V-Notch Impact Test Properties, foot-pounds*				
	RT (80 F)	+10 F	-40 F	-110 F	-320 F
H 925	30	16	9	5.5	3.5
H 1025	75	58	40	15	4.5
H 1150	95	93	76	48	6.5

* Each value is the average of 6 tests.

TABLE XI. ROOM-TEMPERATURE TENSILE PROPERTIES OF 17-4 PH, AGED AND UNAGED, OBTAINED BY HOT-STRETCH-FORMING AS INDICATED (Ref. 8)

H 900 properties given for comparison.

Hot Forming Temperature, F	% Stretch in 2"	Aging Temperature (4 hours)	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2", %
500	9.0	900	198.0	202.2	9.0
650	16.0	900	168.7	190.8	9.0
650	20.0	900	171.8	190.5	9.0
650	33.5	None	133.6	162.4	6.0
650	35.0	900	181.4	194.6	7.5
650	35.0	1050	155.0	160.0	7.5
800	22.0	900	166.6	178.4	11.0
800	42.0	None	116.0	155.8	7.0
800	47.5	900	176.0	190.0	8.0
Standard H 900			173.8	195.0	11.0

Residual stresses developed during cold-fabricating operations may be harmful in applications where stress corrosion is a potential problem. The standard aging treatments used for 17-4 PH would tend to relieve such stresses, but recent work (Ref. 12) has indicated that overaging may be necessary to reduce the residual stresses to a safe level. A method was devised to measure stresses remaining in 17-4 PH wire horseshoe specimens after various aging treatments. The specimens were initially stressed before and after aging to simulate residual stresses from cold-fabricating operations. Residual stresses were measured after the specimens had been given various stress-relieving heat treatments. The results obtained were presented in the form of graphs shown in Figures 5, 6, 7. In each test, it was concluded that adequate stress relieving was not attained at aging temperatures below about 1000 F. This indicates that if adequate strength is not obtained at the aging temperature necessary for adequate stress relieving, the part should be solution treated again at 1900 F after fabricating and then aged under the conditions that will give the desired properties.

Stress aging at high temperature is discussed in the section on Alloy 17-7 PH (p114). The results of such a treatment applied to 17-4 PH, H 900 bar stock are shown in Figure 8.

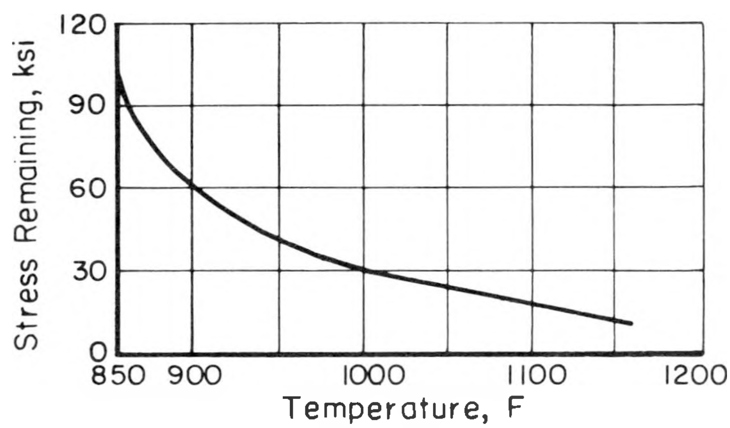

FIGURE 5. RESIDUAL STRESSES IN 17-4 PH STRESSED WIRE SAMPLES AFTER 4-HOUR AGING AT THE INDICATED TEMPERATURES (Ref. 12)

Annealed wires initially stressed to 107,000 psi (equal to 0.2% offset yield strength of annealed 17-4 PH)

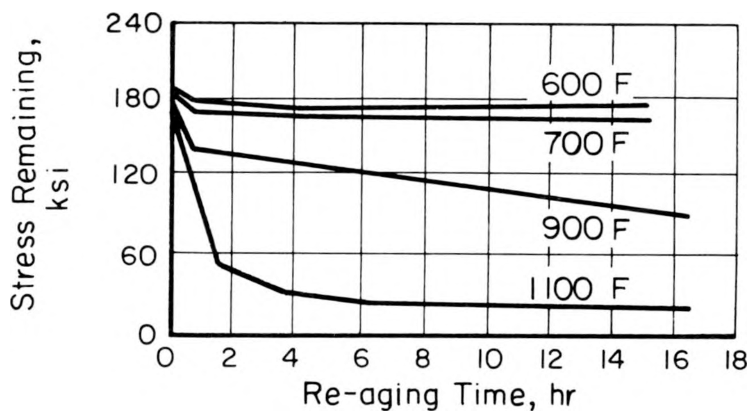

FIGURE 6. RESIDUAL STRESSES IN AGED AND UNAGED STRESSED 17-4 PH WIRES AFTER REAGING AT VARIOUS TEMPERATURES (Ref. 12)

Specimens stressed to 185,000 psi after having been initially at 900 F for 1 hour

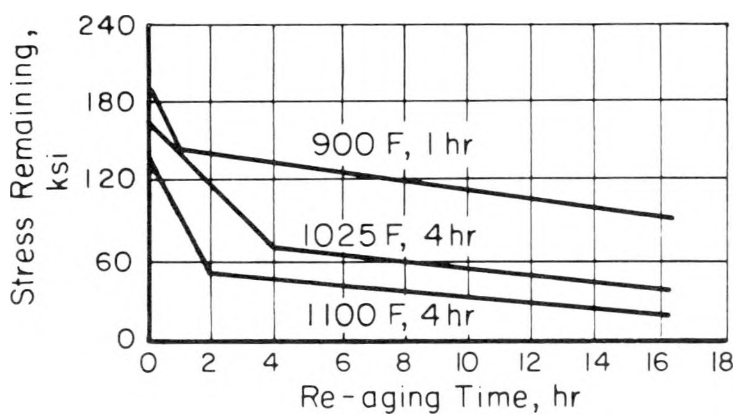

FIGURE 7. RESIDUAL STRESSES IN AGED AND STRESSED 17-4 PH WIRES AFTER VARIOUS REAGING TREATMENTS (Ref. 12)

Specimens were initially aged as indicated, stressed to the 0.2% offset yield strength typical for each aging temperature and then reaged at the same temperatures for various times.

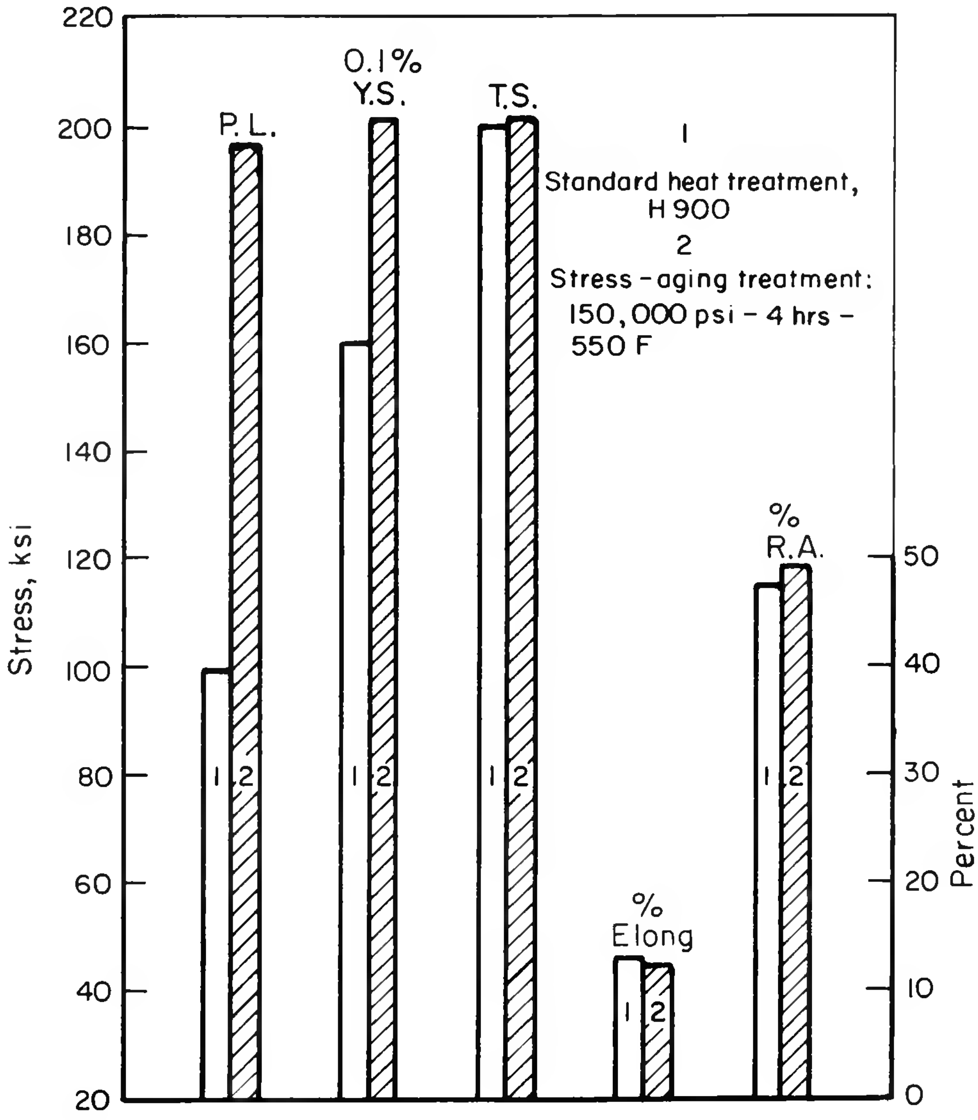

FIGURE 8. EFFECT OF A STRESS-AGING TREATMENT APPLIED TO 17-4 PH, H-900 BAR STOCK ON TENSILE PROPERTIES (Ref. 13)

15-5 PH. This new martensitic PH stainless steel is a modification of 17-4 PH, having a lower chromium and slightly higher nickel content. This composition is designed to be essentially free of delta ferrite, and therefore not subject to the variations in transverse ductility discussed in the section on 17-4 PH.

From a heat treatment standpoint, 15-5 PH is similar to 17-4 PH. It is solution treated at 1900 $\pm$ 25 F for one-half hour, oil or air cooled, and then aged at temperatures from 900 F to 1150 F, depending on the properties desired. The standard treatments are shown in Table XII. The mechanical properties developed by the various aging treatments are summarized in Table XIII. These properties are in the longitudinal direction on specimens taken from an intermediate location within the bar, and from either air-melted or consumable-electrode vacuum-arc remelted stock. The results show that, under the conditions specified, the strength and ductility of 15-5 PH are identical to those of 17-4 PH, and vary with heat treatment in the same manner. The advantage of the 15-5 PH, however, lies in the improved properties in the short transverse direction. This is brought about by the elimination of delta ferrite in the structure of 15-5 PH. In spite of this however, the short transverse ductility in the center of a bar (air melted) is definitely inferior to that at intermediate and edge locations (Ref. 10). The difference is the result of segregation in the center during solidification of the ingot. The segregation may be eliminated by vacuum-arc remelting resulting in greatly improved short transverse ductility across the center of large shapes. The mechanical properties of vacuum-arc remelted bar in the transverse direction at intermediate and center locations have been determined, and the results given in Table XIV show the effect of aging conditions on such properties. The H 900

TABLE XII. STANDARD HEAT TREATMENTS FOR 15-5 PH STAINLESS STEEL (Ref. 14)

Condition A

Solution Treated
1900 F ± 25 F 1/2 Hr
Oil or Air Cool

Condition	Heat to ± 15 F	Hold for, Hours	Cool
H 900	900 F	1	Air
H 925	925 F	4	Air
H 1025	1025 F	4	Air
H 1075	1075 F	4	Air
H 1100	1100 F	4	Air
H 1150	1150 F	4	Air
H 1150-M	1400 F	2	Air
(Double Overaged)		Followed by	
	1150 F	4	Air

TABLE XIII. EFFECT OF AGING CONDITIONS ON TYPICAL MECHANICAL PROPERTIES OF 15-5 PH BAR STOCK (Ref. 14)

Longitudinal Direction - Intermediate Location
Air Melted or Consumable Electrode Vacuum Arc Remelted

	Condition						
	H 900	H 925	H 1025	H 1075	H 1100	H 1150	H 1150-M
Ultimate Tensile Strength, ksi	200	190	170	165	150	145	125
0.2% Yield Strength, ksi	185	175	165	150	135	125	85
Elongation, % in 2" or 4 x D	14.0	14.0	15.0	16.0	17.0	19.0	22.0
Reduction of Area, %	50.0	54.0	56.0	58.0	58.0	60.0	68.0
Hardness, Rockwell	C 44	C 42	C 38	C 36	C 34	C 33	C 27
Impact, Charpy V-Notch, ft-lbs	20	25	35	40	45	50	100

TABLE XIV. EFFECT OF AGING CONDITIONS ON TYPICAL MECHANICAL PROPERTIES OF VACUUM ARC REMELTED 15-5 PH BAR STOCK (Ref. 14)

Transverse Direction - Intermediate and Center Location

	H 900		H 925		H 1025		H 1075		H 1100		H 1150		H 1150-M	
	I*	C*	I	C	I	C	I	C	I	C	I	C	I	C
Ultimate Tensile Strength, ksi	200	200	190	190	170	170	165	165	150	150	145	145	125	125
0.2% Yield Strength, ksi	185	185	175	175	165	165	150	150	135	135	125	125	85	85
Elongation, % in 2" or 4 x D	10.0	10.0	11.0	11.0	12.0	12.0	13.0	13.0	14.0	14.0	15.0	15.0	18.0	18.0
Reduction of Area, %	30.0	30.0	35.0	35.0	42.0	42.0	43.0	43.0	44.0	44.0	45.0	45.0	50.0	50.0
Hardness, Rockwell	C 44	C 44	C 42	C 42	C 38	C 38	C 36	C 36	C 34	C 34	C 33	C 33	C 27	C 27
Impact, Charpy V-Notch, ft-lbs														
Notch Axis Longitudinal	7		17		27		30		30		50		100	
Notch Axis Transverse	8		12		25		25		25		45		70	

* I - Intermediate Location
C - Center Location

condition results in maximum tensile and yield strength with good transverse ductility. By aging at higher temperatures both ductility and impact strength are improved, with some sacrifice in tensile properties.

PH 13-8 Mo. This alloy is being developed to provide a martensitic stainless steel free of delta ferrite in which aluminum is used as the hardening element. It is being evaluated for parts of large cross section that require good centerline transverse ductility.

Heat Treatment. PH 13-8 Mo is solution annealed at 1700 F for one-half hour to dissolve the hardening element, aluminum, in the austenite matrix, air cooled or oil quenched to transform the metallurgical structure to martensite, then heated four hours at 950 to 1150 F for hardening by precipitation of intermetallic compounds. The alloy responds fully to heat treatment by the braze cycle processes used for semiaustenitic precipitation-hardenable stainless steels. Table XV lists typical room-temperature properties as annealed (Condition A), as aged at 950 F (Condition H 950) after annealing, and on being given two different braze cycle heat treatments (Conditions BCHT 950 and BCHT 1050). Variations of the braze cycle process from the standard heat treatment are: shorter time at the solution-annealing temperature; slower cooling rate to 1000 F; and, refrigeration at -100 F for 8 hours prior to aging.

Figures 9 and 10 show the effect of variation in solution-annealing temperature on the mechanical properties of PH 13-8 Mo subsequently aged at 950 F and 1050 F. These data are from tests on specimens from the short transverse direction. They illustrate what may happen to the strength and toughness of PH 13-8 Mo when brazing temperatures are changed to suit the flow temperature of the brazing alloy or to heat treat more effectively other alloys in a brazed assembly.

TABLE XV. TYPICAL MECHANICAL PROPERTIES OF 3" x 8 " PH 13-8 Mo BILLET AFTER STANDARD AND BRAZE-CYCLE HEAT TREATMENTS(1) (Ref. 15)

Condition	Test Direction(1)	Ultimate Tensile Strength, ksi	0.2% Yield Strength, ksi	Elongation in 4 x D, %	Reduction of Area, %	Impact Charpy V-Notch ft-lbs	Hardness, Rockwell
A	Longitudinal	161	101	15.6	66.6	--	C 35
H 950	Longitudinal	214	188	15.9	63.3	18	C 46
	Transverse	214	189	14.1	63.2	--	
	Short Transverse	217	192	15.0	65.2	15	
BCHT 950*	Longitudinal	226	207	13.1	65.7	22	C 47
	Transverse	223	200	14.8	62.8	--	
	Short Transverse	224	199	17.0	60.5	--	
BCHT 1050*	Longitudinal	212	206	14.1	67.3	96	C 45
	Transverse	209	201	13.1	63.8	--	
	Short Transverse	202	194	15.0	62.3	--	

(1) All test specimens taken from intermediate location in the bar.

* 1700 F for 15 minutes, cool to 1000 F in 1 hour, plus -100 F for 8 hours, 950 F or 1050 F for 1 hour, air cool.

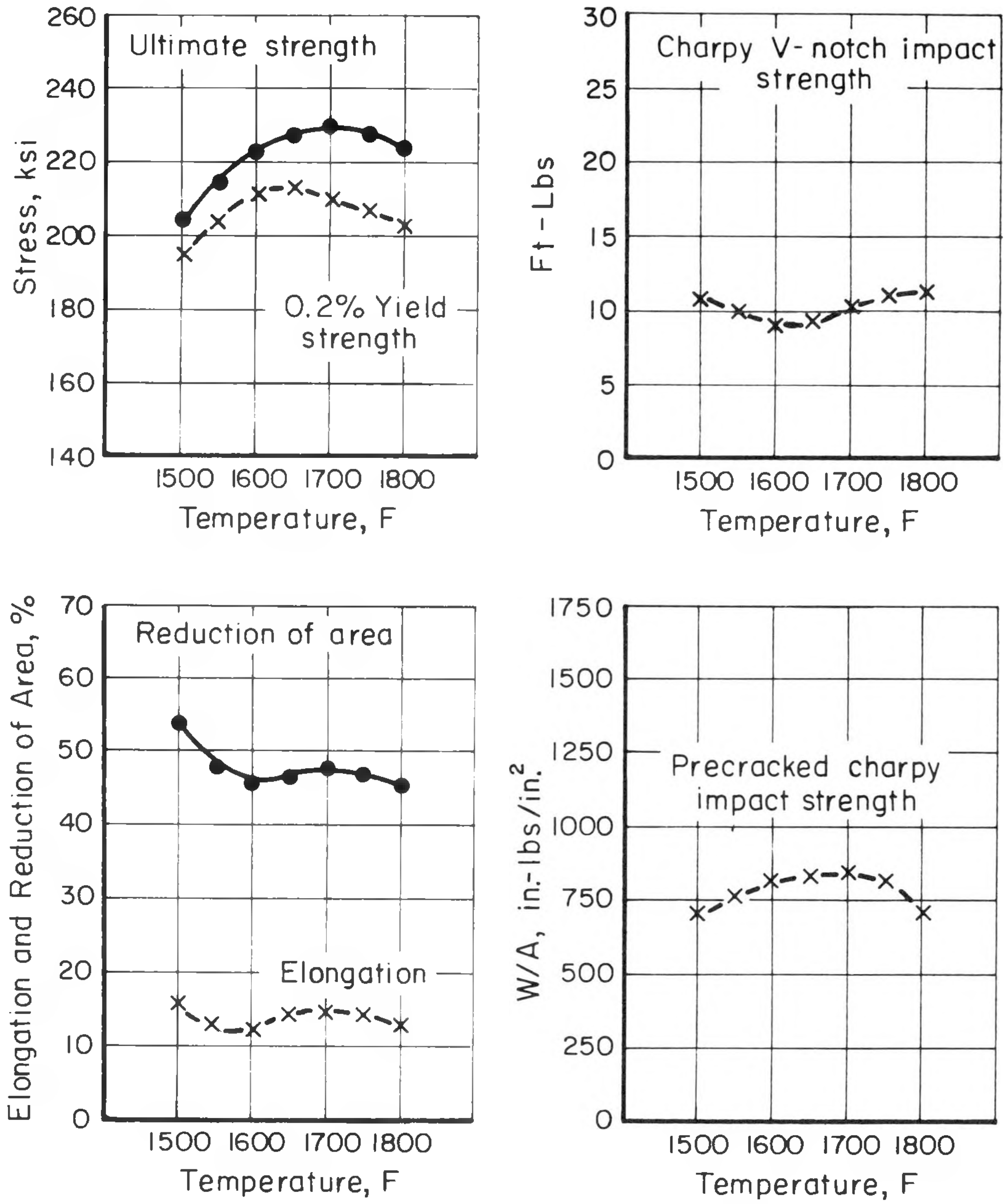

FIGURE 9. EFFECT OF VARIATION IN SOLUTION-ANNEALING TEMPERATURE ON THE MECHANICAL PROPERTIES OF PH 13-8 Mo AGED AT 950 F (Ref. 15)

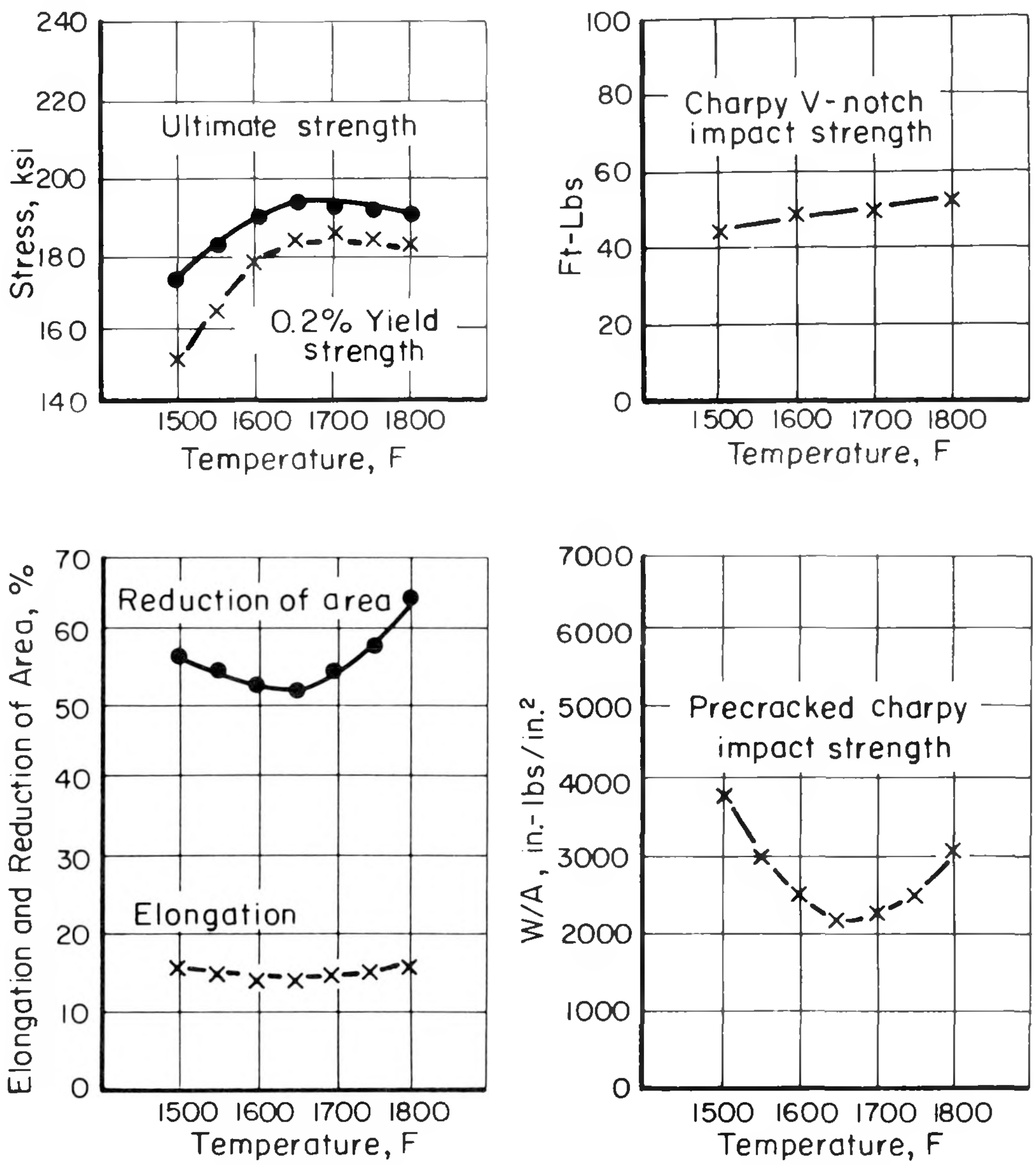

FIGURE 10. EFFECT OF VARIATION IN SOLUTION-ANNEALING TEMPERATURE ON THE MECHANICAL PROPERTIES OF PH 13-8 Mo AGED AT 1050 F (Ref. 15)

Strength is lowered and ductility and toughness values are increased when aging temperature exceeds 950 F. The effects of aging four hours at temperatures up to 1100 F on the mechanical properties of PH 13-8 Mo are shown in Figure 11.

Hot Working. Because the alloy is still in the development stage, it is suggested that information on hot-working procedures be obtained from the producer.

Cold Working. PH 13-8 Mo is intended primarily for large bar, forging and plate applications and so will not be cold worked.

AM 362. Two thermal processing steps are required to fully heat treat this martensitic alloy. On cooling from the recommended annealing temperature, 1500 F, martensite formation begins at about 500 F and is complete at 320 F. After cooling to room temperature, the alloy is further strengthened by precipitation hardening at temperatures from 900 F to 1150 F for various times. Figure 12 diamgrams a complete heat-treating cycle and shows a typical heating and cooling cycle, the phase changes that take place, and the condition of the alloy after each step in the cycle. Typical properties of AM 362 in the standard annealed and aged conditions shown in Figure 12 are given in Table XVI.

Variation in Aging Treatment. AM 362 is hardened during aging by precipitation of titanium-containing intermetallic compounds. The precipitation reaction is typical in that highest strengths are achieved on long exposure at the lower temperatures, while increased aging temperature for shorter times improves ductility and toughness at some sacrifice of strength. The effects of variations in aging temperature and time on strength, ductility and impact properties are shown in Table XVII.

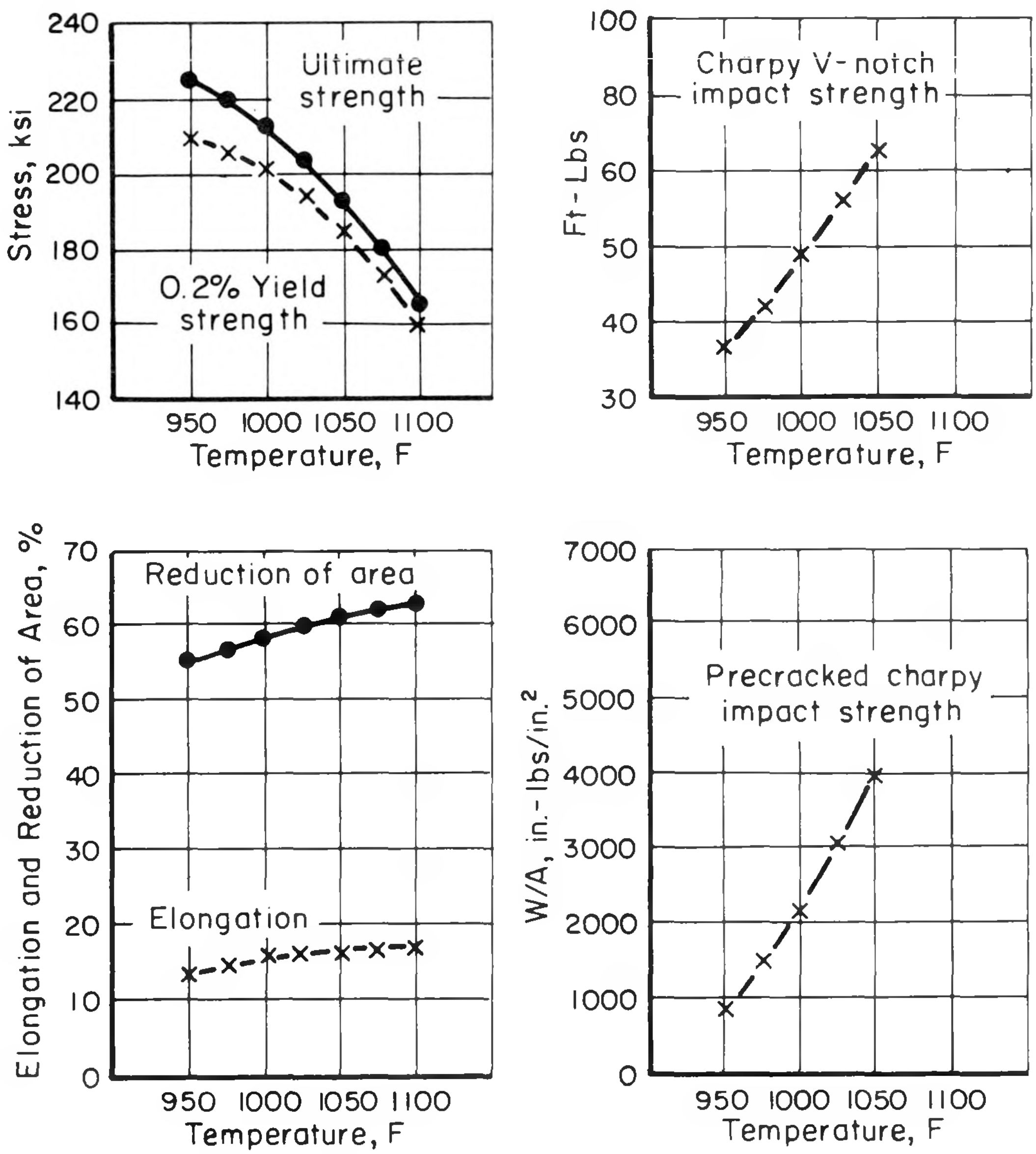

FIGURE 11. EFFECT OF VARIATION IN AGING TEMPERATURE ON THE MECHANICAL PROPERTIES OF PH 13-8 Mo (Ref. 15)

Solution annealed at 1700 F and aged for 4 hours.

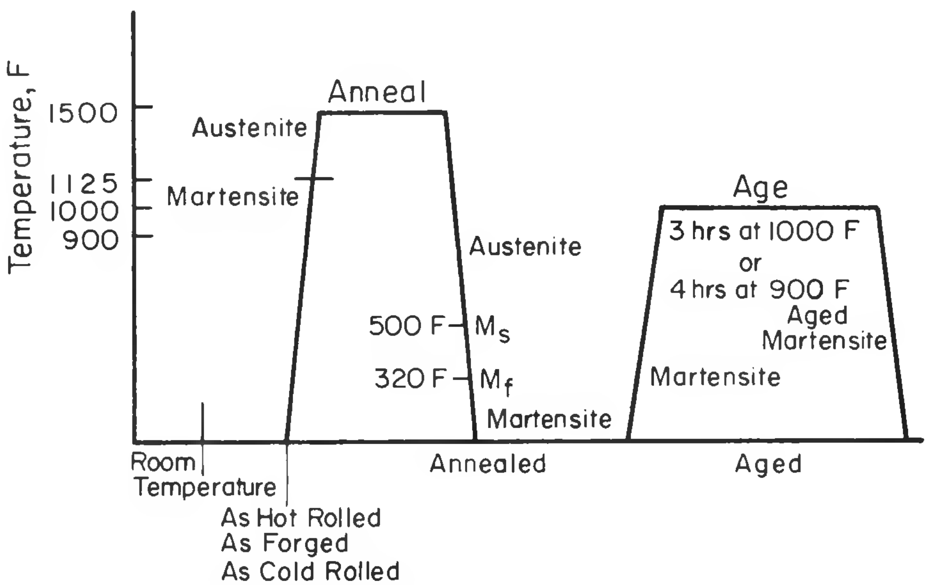

FIGURE 12. EFFECT OF THERMAL PROCESSING ON THE METALLURGICAL STRUCTURE AND CONDITION OF AM-362 STEEL (Ref. 16)

TABLE XVI. TYPICAL MECHANICAL PROPERTIES OF PH 13-8 Mo OBTAINED BY STANDARD HEAT TREATMENTS (Ref. 16)

See Figure 12

Condition	Ultimate Tensile Strength, ksi	Yield Strength, 0.2% Offset, ksi	Elongation in 2 inches, %	Hardness, Rockwell C	Charpy V-Notch Strength, ft-lbs
Annealed at 1500 F, 1 hr per inch of thickness	130	100	15	25	-
Aged, 3 hours, 1000 F	160	150	-	37	30
Aged, 4 hours, 900 F	190	180	-	42	5

TABLE XVII. EFFECTS OF AGING TEMPERATURE AND TIME ON THE MECHANICAL PROPERTIES OF AM 362 BAR (Ref. 16)

Aging Treatment (plus air cool) F	hr	Hardness, Rockwell C	Yield Strength, 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in., %	Reduction in Area, %	Impact Strength, ft-lb
900	8	41	182	188	13	54	6
950	4	39	172	177	14	57	10
975	4	38	167	172	15	58	15
1000	3	37	160	165	16	60	30
11050	2	33	144	152	18	64	50
1150	1	30	115	140	21	68	80
Annealed		25	108	125	16	68	

Hot Working. Initial hot-working temperatures for forging and rolling should be in the range of 2000 to 2250 F. Finishing temperatures can be as low as 1400 to 1500 F because AM 362 does not display hot shortness.

Cold Working. AM 362 has a low work hardening rate. In contrast to some precipitation-hardenable stainless steels of this type, the untempered martensitic structure of the annealed material can be cold worked severely without intermediate annealing. For example, tubing and wire have been reduced over 90 percent by drawing without reannealing. Significant increases over annealed and aged strengths can be obtained by aging after cold working. Examples of strengths exhibited by tubing cold reduced by drawing and subsequently aged at 900 F are shown in Table XVIII. The mechanical properties of wire, cold drawn and aged at temperatures from 800 to 950 F are listed in Table XIX.

AM 363. This alloy is a low carbon martensitic stainless steel produced primarily as strip and plate. It is annealed by heating at 1500 - 1600 F for short times and air cooling or quenching to room temperature. On cooling, austenite transforms to martensite over the temperature range from 770 to 500 F. The mechanical properties of material in the annealed condition are; 100,000 psi yield strength, 120,000 psi ultimate strength, and 10 percent elongation.

AM 363 has low total alloy content. It exhibits a weak precipitation-hardening response, from small amounts of nickel and titanium in the composition, on reheating at 900 to 1100 F. Yield strength can be increased about 10 percent by this aging treatment.

TABLE XVIII. EFFECT OF COLD-DRAWING AND AGING ON THE MECHANICAL PROPERTIES OF AM 362 TUBING (Ref. 17)

Diameter, inch	Thickness, inch	Condition	Yield Strength, 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in., %
1.750	0.200	As hot extruded + 1500 F, 1 hour, AC + 900 F, 8 hours, AC	172	180	15
1.615	0.154	Cold drawn 27.4% + 900 F, 8 hours, AC	185	186	8
1.350	0.094	Cold drawn 62% + 900 F, 8 hours, AC	187	200	7
1.000	0.059	Cold drawn 70% + 900 F, 8 hours, AC		205	10
0.987	0.041	Cold drawn 87% + 900 F, 8 hours, AC	206	206	11

TABLE XIX. EFFECT OF COLD-DRAWING AND AGING ON THE MECHANICAL PROPERTIES OF AM 362 WIRE (Ref. 17)

Wire Diameter, inch	Cold Reduction, %	Aging Treatment F	Aging Treatment hour	Yield Strength, 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Reduction in Area, %
0.111	88	None		165	185	20
		800	4	210	225	25
		850	4	215	230	30
		900	4	210	225	45
		950	4	205	220	40
0.083	10	None		131	136	70
		900	4	192	194	58
	75	900	4	196	202	47
	89	900	4	201	205	50
0.009	80	None		210	235	
		800	1	256	257	
		850	1	250	251	
		900	1	250	250	
		950	1	240	240	

Cold Working. AM 363 may be formed in either the annealed or aged conditions. The alloy has good bend properties and may be formed readily by brake forming. However, fabrication by stretch wrapping or deep drawing is not recommended because low uniform elongation values will cause localized necking or failure. Elevated-temperature forming is not feasible because elongation values decrease with increasing temperature as shown in Table XX.

Custom 455. This martensitic alloy is available as billet, bar, wire and strip. It has a low work hardening rate and can be formed in the annealed condition. Cold work prior to aging results in higher strengths than those obtained by heat treatment only. In the precipitation-hardened condition it can be used for service temperatures up to 800 F.

Thermal Treatments. Custom 455 is annealed by heating at 1500 to 1550 F for one-half hour and water quenching. At the annealing temperature the hardening elements, copper and titanium, are taken into solution in the austenite matrix. Water quenching transforms the austenite to low strength, low carbon martensite. To improve transverse properties in thick sections or to obtain lower hardness as annealed, Custom 455 may be water quenched from 1800 F before annealing at 1500 F.

The alloy is precipitation hardened by heating at 900 to 1000 F for four hours. The effect of aging temperature on room-temperature tensile properties is shown in Figure 13. Aging at 950 F results in the best combination of properties. Lower temperatures may be used to obtain higher strength. Ductility may be improved by employing higher aging

TABLE XX. EFFECT OF TEMPERATURE ON THE MECHANICAL PROPERTIES OF ANNEALED AND AGED AM 363 STRIP (Ref. 18)

Test Temperature, F	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Notched[(a)] Tensile Strength, ksi	Notched/ Unnotched Ultimate Strength Ratio	Yield Strength/ Ultimate Tensile Strength Ratio
A. Longitudinal Specimens						
-100	123	143	17.0	152	1.06	0.86
- 50	118	135	15.0	147	1.09	0.87
75	107	123	12.5	135	1.15	0.87
200	104	115	12.0	129	1.12	0.90
400	93	105	8.5	119	1.13	0.89
600	90	98	5.0	109	1.10	0.92
B. Transverse Specimens						
-100	130	147	10.0	160	1.09	0.88
- 50	124	134	11.5	151	1.13	0.93
75	112	125	11.5	142	1.14	0.90
200	109	118	9.5	134	1.14	0.92
400	101	108	6.0	122	1.13	0.94
600	96	103	3.0	109	1.06	0.93

(a) 0.0007-inch notch radius, K_t = 15 to 18

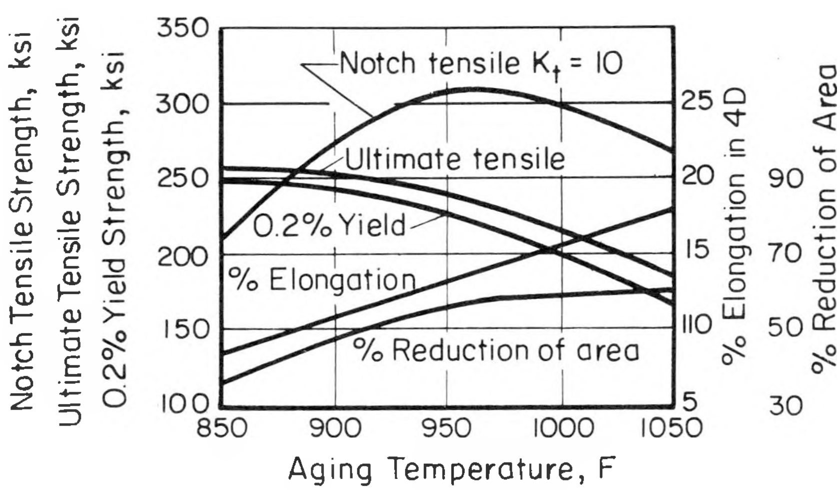

FIGURE 13. EFFECT OF AGING TEMPERATURE ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF CUSTOM 455 (Ref. 19)

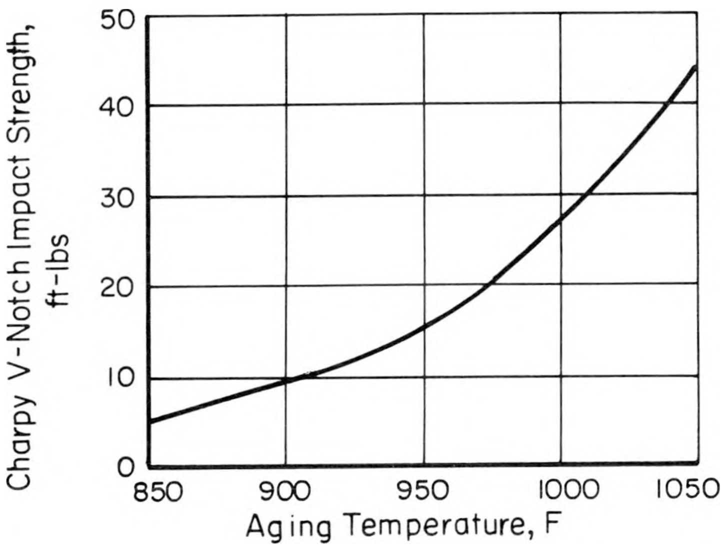

FIGURE 14. EFFECT OF AGING TEMPERATURE ON THE ROOM-TEMPERATURE CHARPY V-NOTCH STRENGTH OF CUSTOM 455 (Ref. 19)

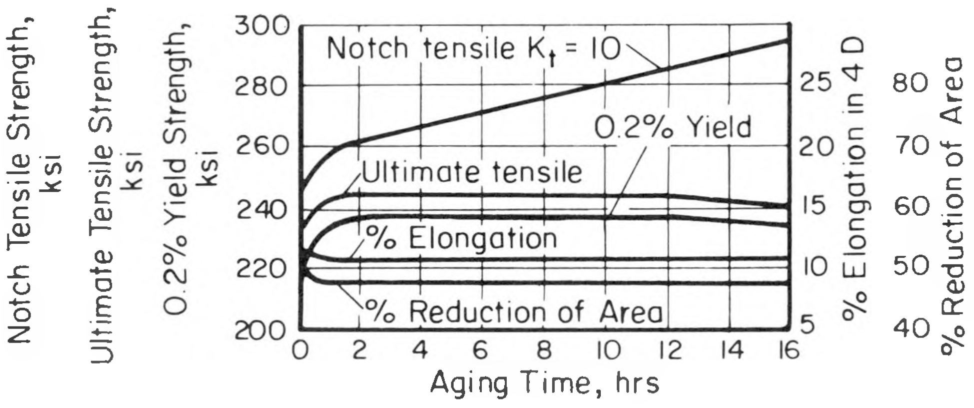

FIGURE 15. EFFECT OF AGING TIME AT 900 F ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF CUSTOM 455 (Ref. 19)

temperatures. The effect of aging temperature on impact strength is shown in Figure 14. The effect of aging time at 900 F on room-temperature properties is shown in Figure 15. Although precipitation hardening is essentially complete after two hours at temperature, aging for four hours is recommended to improve notch properties.

Hot Working. Custom 455 should be heated uniformly at 1900 F to 2100 F for the beginning of hot work. Finishing temperature range should be 1500 to 1700 F to produce proper grain size and to obtain satisfactory mechanical properties after heat treatment.

Cold Working. The soft martensite structure of annealed Custom 455 has a low work hardening rate. It can undergo considerable cold work before reannealing is necessary to restore ductility. Table XXI and Figure 16 show the effects of increasing reductions by cold drawing on the mechanical properties of 1/4-inch-diameter wire. Further increases in strength are obtained by aging after cold working. This is shown by comparison of Table XXII and Figure 17 with Table XXI and Figure 16, respectively.

AFC-77. This alloy was developed under Air Force Contract to provide a very high-strength, corrosion-resistant material for use at temperatures as high as 1200 F. It differs from other precipitation-hardenable stainless steels in that nickel is replaced by cobalt in the composition as an austenite promoter and the fact that hardening results from two precipitation reactions. The alloy is available in the form of bar, billet and sheet.

TABLE XXI. EFFECT OF COLD WORK ON COLD-DRAWN 1/4-INCH-DIAMETER CUSTOM 455 WIRE (Ref. 19)

Cold Work, %	Yield Strength, 0.2% ksi	Ultimate Tensile Strength, ksi	Elongation in 2", %	Reduction of Area, %	Hardness, Rockwell C
0	122	148	13	68	31/32
10	144	156	10	65	32
20	154	164	9	62	33
30	161	172	8	60	34
40	172	180	7	58	35
50	179	187	6	57	36
60	189	195	5	54	37

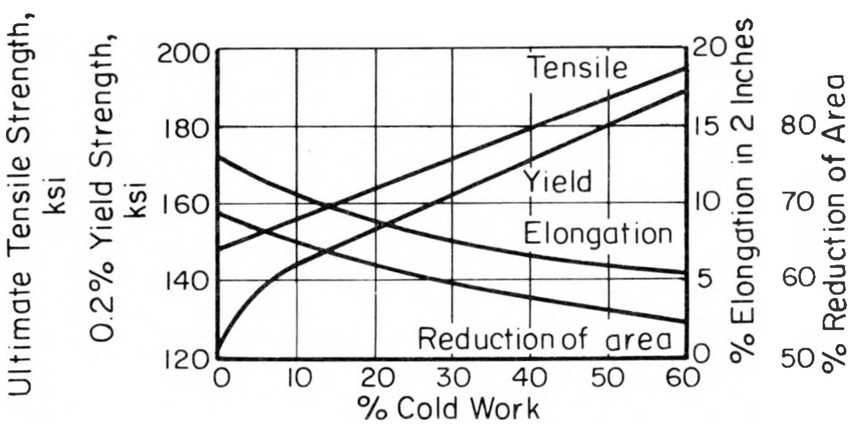

FIGURE 16. EFFECTS OF COLD WORK ON COLD-DRAWN 1/4-INCH-DIAMETER CUSTOM 455 WIRE (Ref. 19)

TABLE XXII. EFFECT OF AGING (4 HOURS, 900 F) ON THE MECHANICAL PROPERTIES OF COLD-DRAWN 1/4-INCH-DIAMETER CUSTOM 455 WIRE (Ref. 19)

Cold Work Prior to Aging	Yield Strength, 0.2%, ksi	Ultimate Tensile Strength, ksi	Elongation in 2", %	Hardness, Rockwell C
0	228	248	6	49
10	244	258	5	49/50
20	252	260	5	50
30	256	264	4	51
40	260	268	4	52
50	264	274	4	52/53
60	268	280	4	53

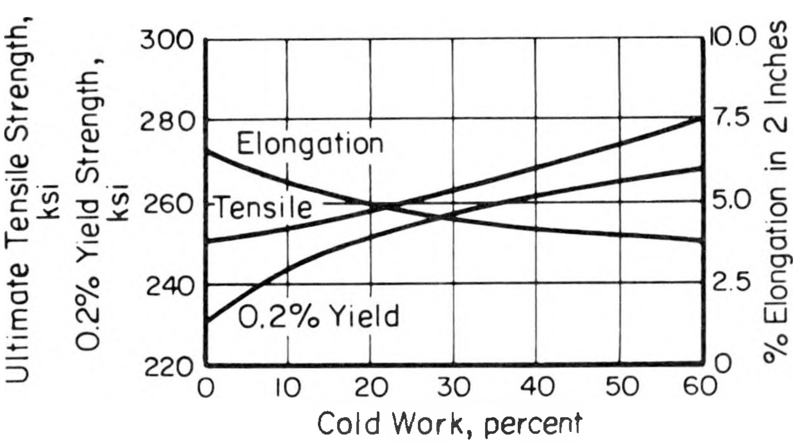

FIGURE 17. EFFECT OF AGING (4 hours, 900 F) ON THE MECHANICAL PROPERTIES OF COLD-DRAWN 1/4-INCH-DIAMETER CUSTOM 455 WIRE (Ref. 19)

Heat Treatment. AFC-77 may be solution annealed at temperatures from 1800 to 2000 F. It is oil quenched from these temperatures and then, although classified as a martensitic alloy, refrigerated at -100 F for one-half hour to promote the austenite-to-martensite transformation.

The alloy is then double aged two plus two hours at either 900 or 1100 F to develop high strength and to reduce the amount of retained austenite in the structure to the lowest possible level.

The alloy is chemically balanced to have an austenitic structure free of delta ferrite at the normal annealing temperatures. In the fully heat-treated condition the structure is essentially martensite, but may have 2 to 5 percent retained austenite. During aging, two precipitation reactions take place. Up to about 900 F, hardening is related to the precipitation of carbides in the martensite. Above 1000 F, precipitation of Fe_2Mo Laves phase and the chi phase has been identified as the primary strengthening mechanism.

A large number of heat-treating sequences can be used because of the wide temperature ranges permissible for both annealing and aging. Figure 18 shows the effects of variation in annealing temperature on the mechanical properties of AFC-77 after hardening at two suggested aging temperatures, 900 F and 1100 F. Figure 19 illustrates the effect of variation in aging temperature on mechanical properties after solution annealing at 1800, 1900, and 2000 F. These curves show that increased aging temperatures up to 900 F raise both yield and ultimate strength. At aging temperatures from 900 to 1100 F, yield strengths are lowered while ultimate strengths generally are increased. No conclusive explanation has been found for the loss of yield strength.

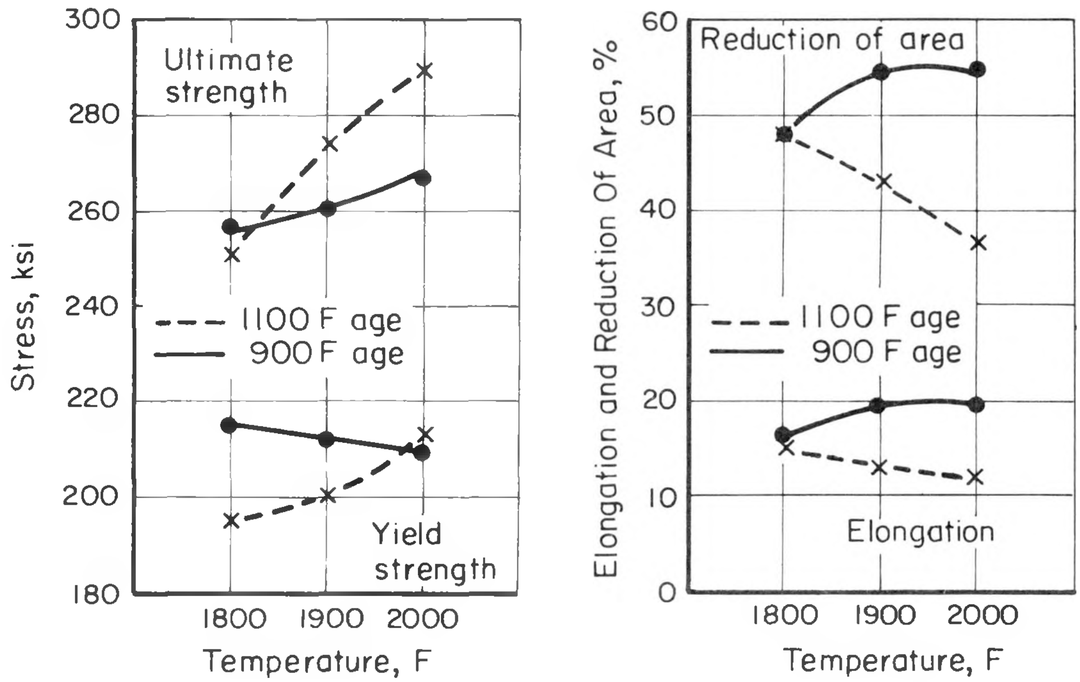

FIGURE 18. EFFECT OF VARIATION IN ANNEALING TEMPERATURE ON THE MECHANICAL PROPERTIES OF AFC-77 AGED AT 900 AND 1100 F (Refs. 20, 21)

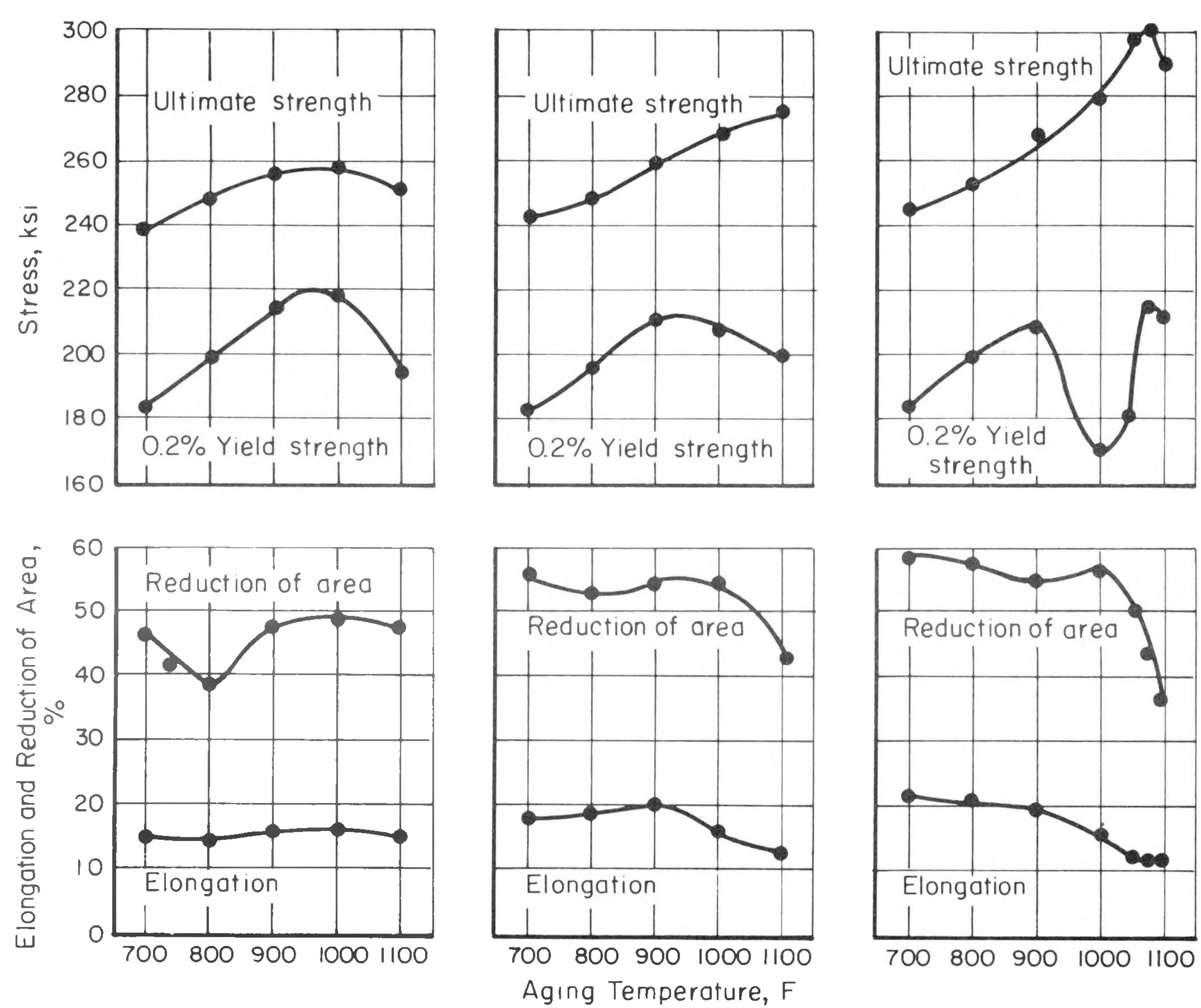

FIGURE 19. EFFECT OF VARIATION IN AGING TEMPERATURE ON THE MECHANICAL PROPERTIES OF AFC-77 AFTER ANNEALING AT 1800 F, 1900 F, 2000 F (Refs. 20, 21)

Hot Working. AFC-77 can be hot worked in the temperature range of 2300 to 1600 F. Cracking may occur at higher temperatures if the composition of the material is such that delta ferrite is formed. Because heavy or fast reductions can result in an increase in temperature of the metal, the initial hot-working temperature should be about 2200 F. Final reductions of 20 to 30 percent at finishing temperatures around 1800 F are recommended to obtain proper grain refinement (Ref. 21).

Cold Working. Strengths higher than those obtained by thermal treatment alone can be obtained when AFC-77 is precipitation hardened after cold reduction in the annealed condition. Table XXIII shows the effect of variations in the amounts of cold work on strength and ductility of sheet material. A comparison with values from the curves in Figure 19 shows that yield strengths are significantly increased by this process at some sacrifice in ductility.

AFC-77 can be formed at room temperature by a number of conventional fabrication methods. The success of the forming operations depends to a large degree on the condition of the material. It has been found that sheet solution annealed at 1900 to 2000 F, oil quenched and aged four hours at 500 F possesses the optimum combination of ductility and strength required for formability. Table XXIV lists the mechanical properties obtained when this heat treatment is used. The formed parts need not be reannealed and can be hardened to the strength level desired by double aging for two plus two hours at the appropriate temperature.

TABLE XXIII. EFFECT OF COLD ROLLING ON THE TENSILE PROPERTIES OF AFC-77 SHEET(1)
(Refs. 20, 21)

Cold Reduction, %	Aging Temperature, F	0.2 Percent Offset Yield Strength, ksi	Ultimate Strength, ksi	Elongation in 2 in, %
5	700	233	247	5
8	700	248	259	4
5	800	272	281	7
10	800	287	297	6
15	800	297	308	5
20	800	301	313	4
30	800	326	331	3
50	800	-	345	0

(1) Processing: Solution annealed at 2000 F for 1/2 hour, oil quenched, cold rolled as indicated and aged 2 plus 2 hours at 700 or 800 F

TABLE XXIV. TENSILE PROPERTIES OF AFC-77 SHEET TEMPERED AT 500 F[a] (Ref. 21)

Heat Treatment	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, in 2 in, %
2000 F 1/2 hr, oil quenched, 500 F 2+ 2 hr	131	211	14
2000 F 1/2 hr, oil quenched, 500 F 4 hr	128	214	15
1900 F 1/2 hr, oil quenched, 500 F 4 hr	140	217	14
1900 F 1/2 hr, oil quenched, 500 F 4 hr	141	218	15

(a) Processed from 10-inch-square ingot to 3/4-inch-thick plate; then rolled to 0.100-inch-thick sheet. All specimens longitudinal direction.

SEMIAUSTENITIC PH STAINLESS STEELS

<u>17-7 PH and PH 15-7 Mo</u>. These steels are essentially austenitic in the annealed or solution-treated condition and are initially hardened by transformation of the austenite to martensite, either by thermal treatment or cold-working. Additional strengthening occurs during aging by the precipitation of hardening phases. 17-7 PH was the first commercial semiaustenitic PH stainless steel, and a considerable background of information on this steel has been accumulated since its introduction in 1948. PH 15-7 Mo is a modification of 17-7 PH in which 2 percent of the chromium is replaced by about 2.25 percent molybdenum. This results in a stronger and more heat resistant alloy. However, the thermal and mechanical treatments to develop properties are essentially the same for the two alloys, and it is desirable to consider them together for the purposes of this memorandum.

The standard treatments for these alloys are illustrated in Figure 20.[a] The chart shows that full hardening may be accomplished by austenite conditioning, transformation and precipitation hardening, or by cold rolling followed only by the age-hardening step. In the annealed condition, A, the alloys are soft and readily formed, but in the cold-rolled Condition C the alloys are strong and have limited fabricability. Typical mechanical properties of the alloys in several standard conditions as well as the properties attained following the various steps in the heat-treating

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(a) This is from Reference 22, the new publication "Armco 17-7 PH and PH 15-7 Mo" available from the Armco Steel Corporation, Middletown, Ohio. The bulletin contains a complete review of the effects of thermal and mechanical treatments on the mechanical properties, as well as much additional information on the physicsl properties, corrosion resistance, welding and fabrication of the alloys.

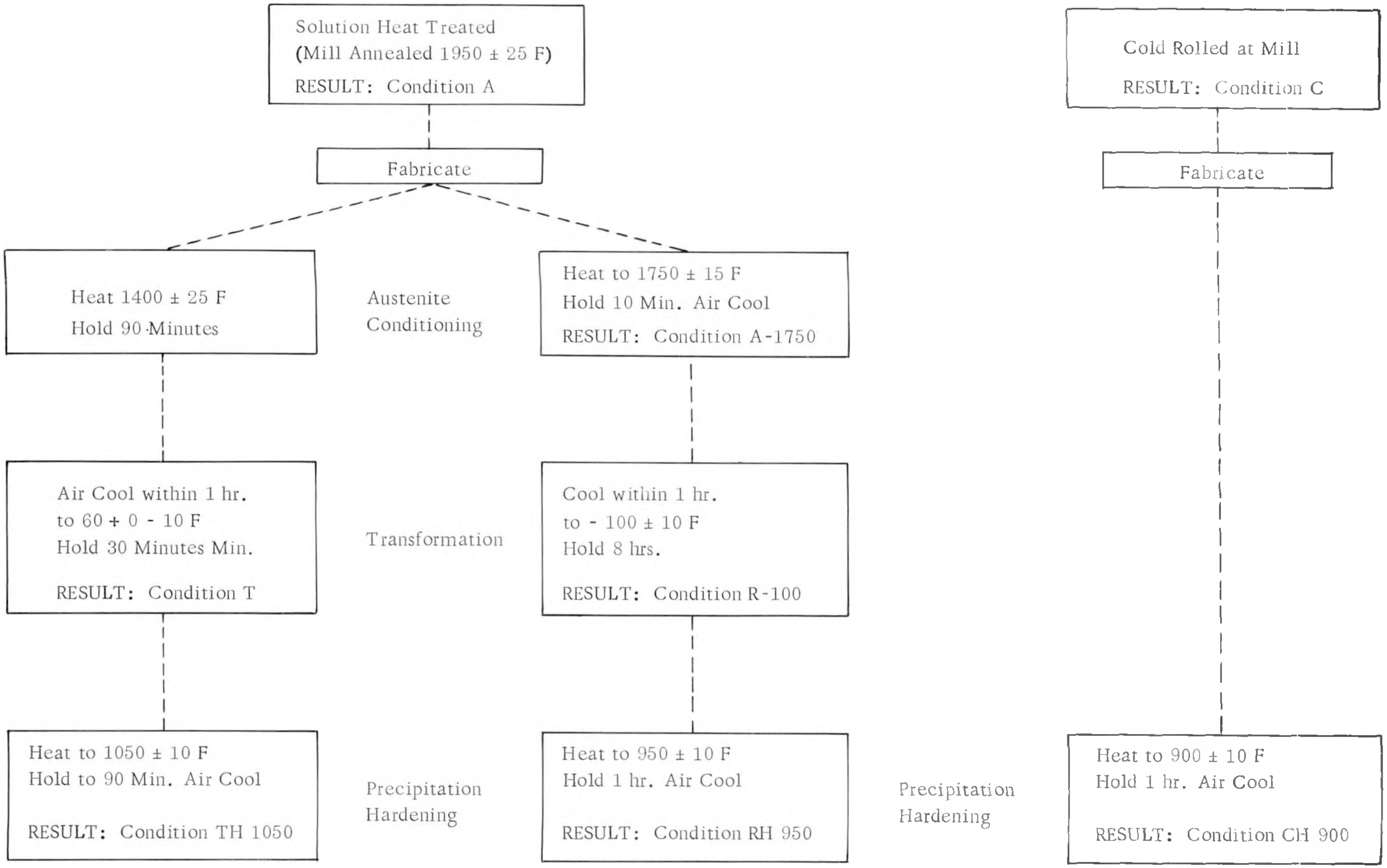


FIGURE 20. STANDARD HEAT TREATMENTS FOR 17-7 PH AND PH 15-7 Mo (Ref. 22)

sequence are given in Table XXV. These are the properties to be expected when using the recommended procedures for each condition. Intentional or unintentional departure from the standard procedures will naturally affect the properties obtained. It is useful to know how the properties vary with variations in the treatments. Such information can be used to explain unusual results obtained in practice, or to determine tolerance limits for specific applications. Sometimes the procedures may be altered intentionally to obtain somewhat different properties.

Variation in Solution Treatment. The recommended annealing or solution-treating temperature is 1950 $\pm$ 25 F with the material being held at temperature for 3 minutes for each 0.1 inch of thickness. With large furnace loads, longer soaking time may be required to assure that the entire load reaches temperature. The effect of variation in the solution-treatment temperature on the properties of as-annealed material is shown in Figure 21. Fairly wide changes in annealing temperature have little effect on the properties in Condition A, but annealing below 1900 F results in somewhat lower elongation. It has been reported (Ref. 1) that annealing above 2000 F causes increased grain size and the formation of larger amounts of delta ferrite and may result in lower ductility. Table XXVI shows that the cooling rate from the annealing temperature may affect the properties. Slow cooling reduces elongation and tends to increase strength. This is probably caused by an adjustment of the austenite composition by precipitation of chromium carbides during the slow cooling of the metal in the austenite-conditioning temperature range. This can then result in some transformation to martensite at room temperature accounting for an increase in strength and loss of dutcility.

TABLE XXV. TYPICAL MECHANICAL PROPERTIES OF 17-7 PH AND PH 15-7 Mo SHEET[1] DEVELOPED BY STANDARD TREATMENTS (Ref. 22)

Condition	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell[2]	Compressive Yield Strength 0.2% Offset, ksi
		17-7 PH			
A	40	130	35	B 85	-
T	100	145	9	C 31	-
TH 1050	185	200	9	C 43	195
A 1750	42	133	19	B 85	-
R 100	115	175	9	C 36	-
RH 950	220	235	6	C 48	227
C	190	220	5	C 43	250
CH 900	260	265	2	C 49	300
		PH 15-7 Mo			
A	55	130	35	B 88	-
T	90	145	7	C 28	-
TH 1050	200	210	7	C 44	217
A 1750	55	150	12	B 85	-
R 100	125	180	7	C 40	-
RH 950	225	240	6	C 48	243
C	190	220	5	C 43	250
CH 900	260	265	2	C 49	300

(1) Transverse tensile properties.
(2) Applies to material 0.010 inch and thicker.

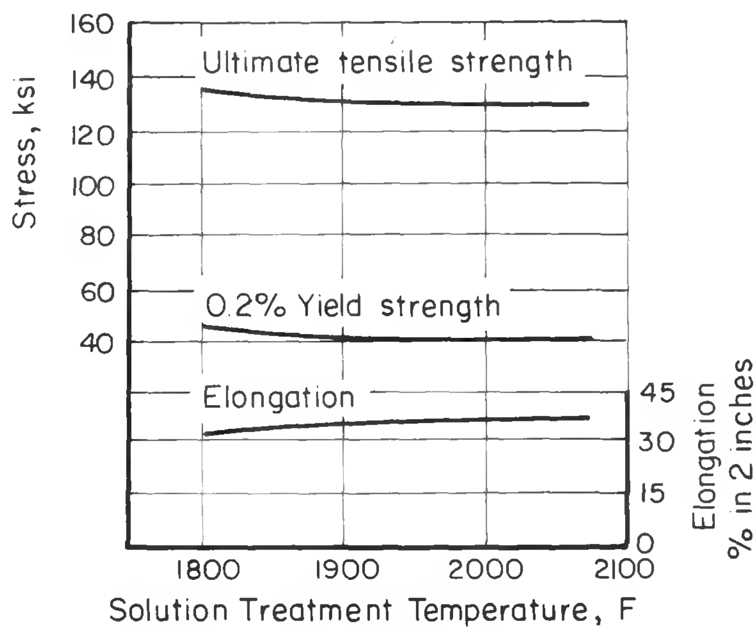


FIGURE 21. EFFECT OF SOLUTION-TREATMENT TEMPERATURE ON THE PROPERTIES OF 17-7 PH AND PH 15-7 Mo IN CONDITION A (Ref. 22)

TABLE XXVI. EFFECT OF SOLUTION-TREATMENT COOLING RATE ON PROPERTIES OF 17-7 PH IN CONDITION A (Ref. 22)

Cooling Time to 1000 F, min	Yield Strength, 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell
Air	39.2	122.6	41.5	B 85
Air	36.9	124.7	39.0	B 84
9.5	38.0	133.4	28.5	B 86
27.0	38.0	135.8	17.7	B 91
110.0	54.6	138.1	12.7	B 100

Therefore, heavy plates may require forced cooling. Cooling is particularly critical on material that is to be given additional cold work.

Variations in annealing temperature have a significant effect on the properties of the 17-7 PH material that is subsequently treated to Conditions TH 1050 or RH 950. This is shown in Figure 22. If the annealing temperature is increased above about 2000 F, the strength of the subsequently hardened material is reduced. The effects of the annealing conditions on PH 15-7 Mo that is later hardened to Conditions TH 1050 and RH 950 are shown in Figure 23. The properties of TH 1050 are practically unaffected by changes in solution-treatment temperature over the range of 1800-2100 F. In the RH 950 condition however, there is a gradual loss in strength of the aged alloy when annealing temperatures above 1950 F are used.

Variation in Conditioning Treatment. Figure 20 shows that the recommended austenite conditioning treatment for the TH 1050 condition is 1400 $\pm$ 25 F for 90 minutes. Figures 24 and 25 show that the properties vary with changes in austenite-conditioning temperature and time. Higher strength is obtained from the 1300 F conditioning temperature but this is accompanied by lower elongation. Severely cold-worked material does not respond well to treatment at 1400 F. A treatment at 1550 F for 90 minutes is recommended in such cases to condition the matrix and to convert to austenite the martensite that was formed by the severe

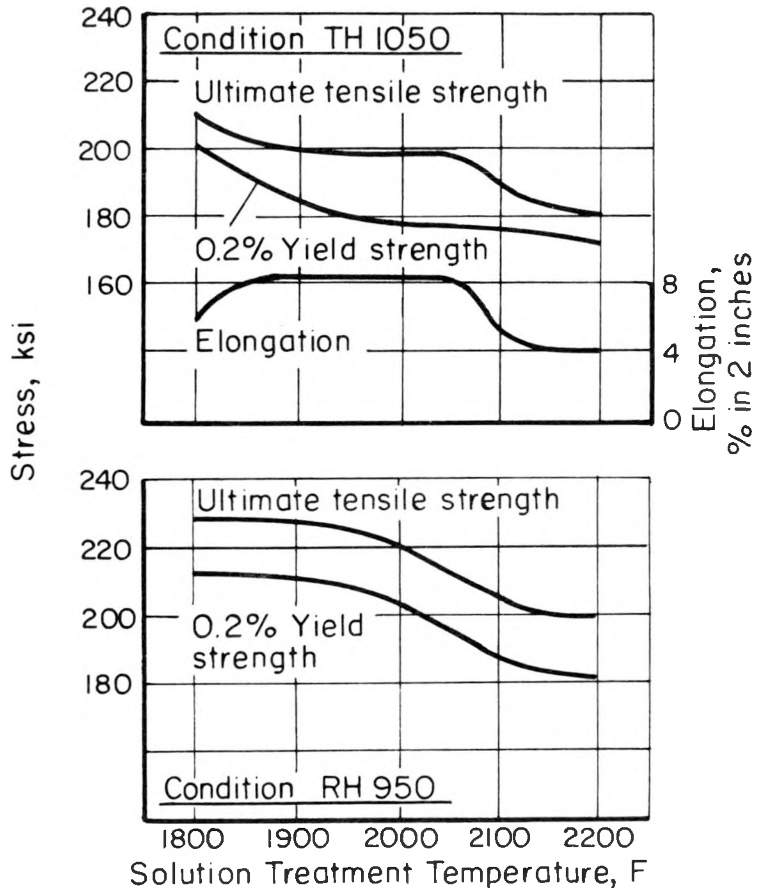


FIGURE 22. EFFECT OF SOLUTION-TREATMENT TEMPERATURE ON THE PROPERTIES OF 17-7 PH IN THE TH 1050 AND RH 950 CONDITIONS (Ref. 22)

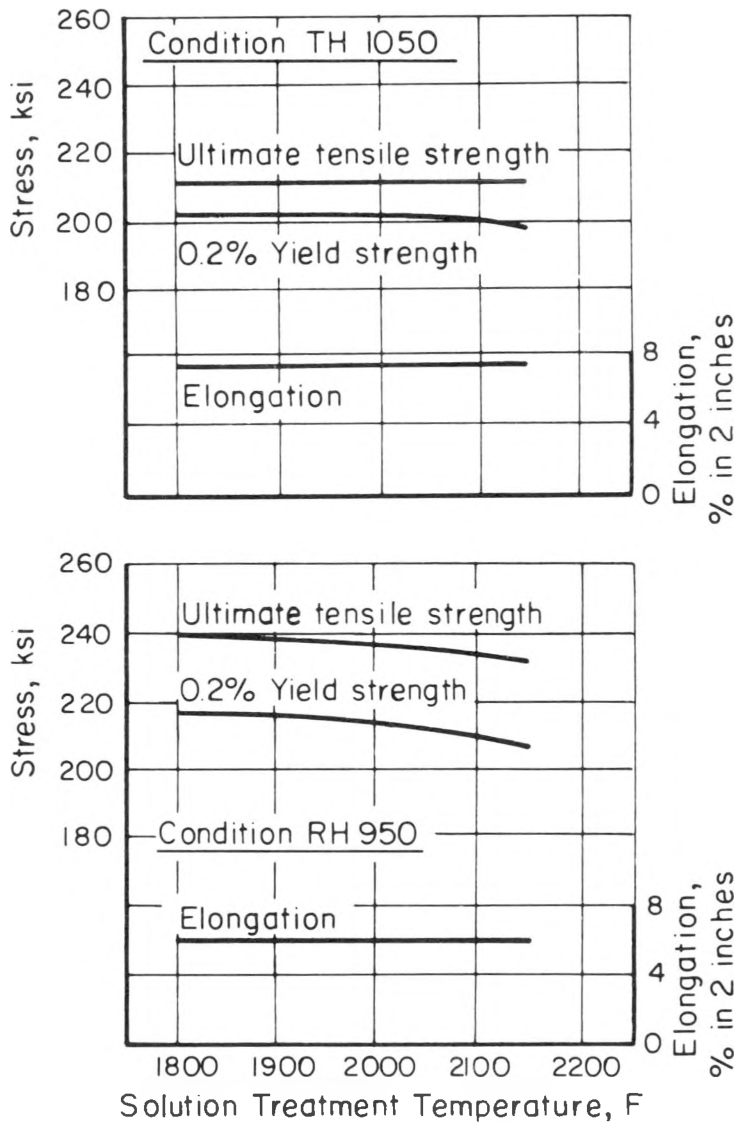


FIGURE 23. EFFECT OF SOLUTION-TREATMENT TEMPERATURE ON THE PROPERTIES OF PH 15-7 Mo IN THE TH 1050 AND RH 950 CONDITIONS (Ref. 22)

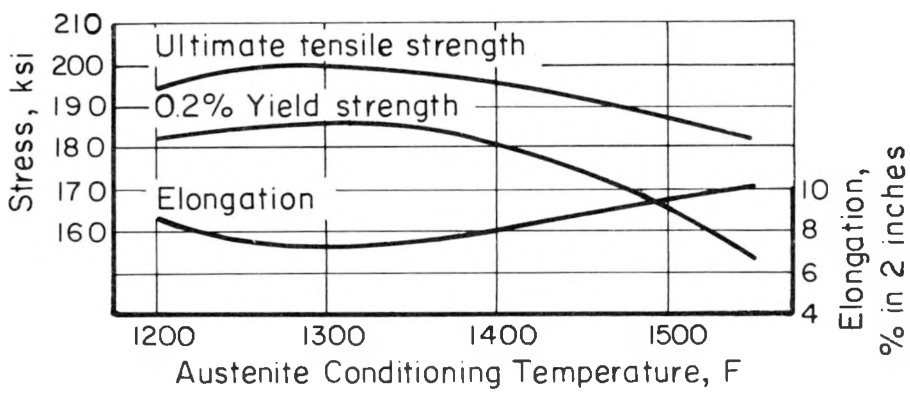


FIGURE 24. EFFECT OF VARIATION IN AUSTENITE CONDITIONING TEMPERATURE ON MECHANICAL PROPERTIES OF 17-7 PH, CONDITION TH 1050 (Ref. 22)

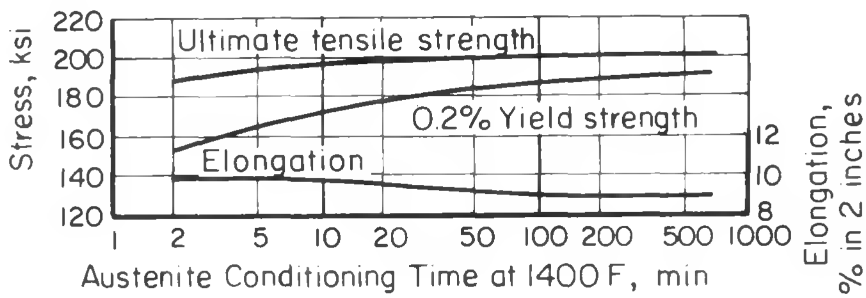


FIGURE 25. EFFECT OF VARIATION IN TIME AT AUSTENITE CONDITIONING TEMPERATURE ON MECHANICAL PROPERTIES OF 17-7 PH CONDITION TH 1050 (Ref. 22)

At 1400 F for time shown, cooled to 60 F, hardened at 1050 F, 90 minutes) cold working. After this treatment the material must be cooled to 0 F to undergo proper transformation (Ref. 2). The longer the holding time at 1400 F, the greater the strength of material subsequently treated to the TH 1050 condition, probably because of more complete carbide precipitation from the austenite. A 90-minute treatment is recommended for better uniformity and adequate strength.

The cooling rate from 1400 F is an important variable because it affects the temperature at which transformation to martensite occurs. The recommended conditions are to cool to 60 F within 60 minutes. If the cooling rate between 200 and 60 F is too low, partial stabilization of austenite can occur and result in less transformation to martensite and consequently lower properties. Prolonged holding times above the M_s but below 1000 F have no effect on mechanical properties. The effect of a delay in cooling from 80 F to 60 F on the yield strength in shown in Figure 26.

If the metal does not reach 60 F within one hour, and stabilization occurs, it may be overcome by cooling to a temperature below 60 F. This is shown in Figure 27. Quenching to 60 F results in highest yield strength but may cause distortion. A process control study on austenite-conditioning temperature and time was reported by Token (Ref. 23). The properties obtained on 17-7 PH sheet, 0.010 to 0.250 inch thick, after austenite conditioning and aging under experimental conditions are summarized in Table XXVII. Although there was a greater deviation in tensile properties related to metal thickness on the specimens treated at 1200 F than on those treated at 1400 F, the author concluded that austenite conditioning at 1200 F will produce acceptable mechanical properties.

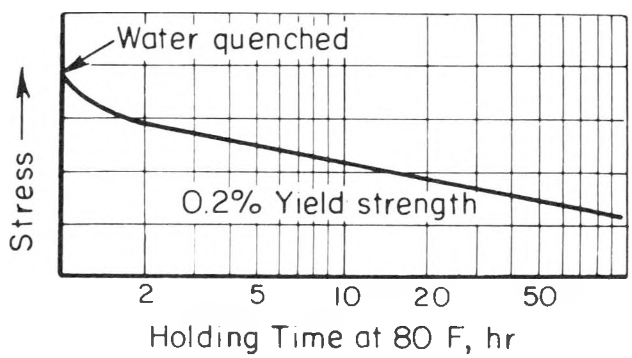


FIGURE 26. EFFECT OF COOLING TIME FROM AUSTENITE CONDITIONING TEMPERATURE TO ROOM TEMPERATURE ON 0.2 PERCENT YIELD STRENGTH OF 17-7 PH, CONDITION TH 1050 (Ref. 22)

1400 F - 90 minutes, air cooled to 80 F, held as indicated above, water quenched to 60 F, hardened 1050 F - 90 minutes

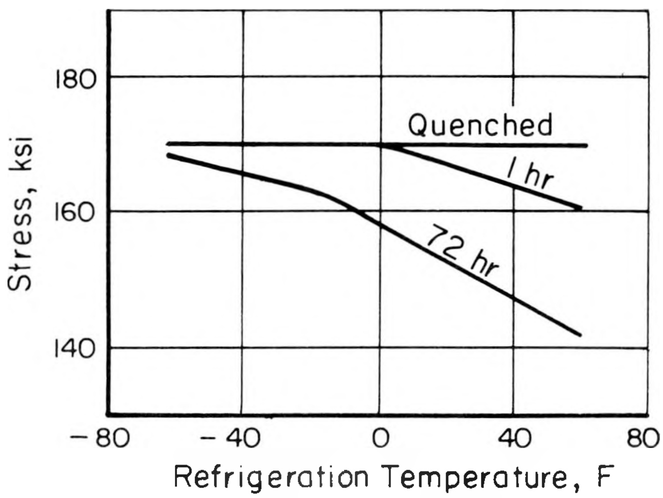


FIGURE 27. EFFECT OF REFRIGERATION TREATMENT DURING TRANSFORMATION ON YIELD STRENGTH OF 17-7 PH, CONDITION TH 1050 (Ref. 22)

1400 F - 90 minutes, air cooled to 80 F for indicated time - cooled to indicated temperature, hardened at 1050 F - 90 minutes

TABLE XXVII. MECHANICAL PROPERTIES OF 17-7 PH SHEET OBTAINED BY VARIATION IN CONDITIONING AND AGING TREATMENTS (Ref. 23)

A 1400 F/1-1/2 hours/AC, -30 F/1 hour, 1050 F/1-1/2 hours
B 1200 F/3/4 hour/AC, - 30 F/1 hour, 900 F/3/4 hour
C 1200 F/1-1/2 hours/AC, -30 F/1 hour, 900 F/1-1/2 hours
D 1200 F/3/4 hour/AC, -30 F/1 hour, 1050 F/3/4 hour
E 1200 F/1-1/2 hours/AC, -30 F/1 hour, 1050 F/1-1/2 hours

Results are average of 5 tests for each treatment

Sheet Thickness, inch	Treatment	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %
0.010	A	191.3	198.0	5.3
0.050	A	188.7	199.1	6.5
0.100	A	192.0	199.7	6.8
0.150	A	189.4	202.4	8.8
0.250	A	182.2	192.4	8.9
0.010	B	198.6	220.9	6.0
0.050	B	185.6	207.8	7.9
0.100	B	197.9	213.4	7.8
0.150	B	182.6	204.0	10.2
0.250	B	182.2	204.9	10.5
0.010	C	213.9	228.6	4.8
0.050	C	203.0	223.0	6.4
0.100	C	206.9	221.1	6.8
0.150	C	197.4	216.5	7.9
0.250	C	198.2	219.1	8.5
0.010	D	197.2	208.6	4.9
0.050	D	182.5	199.3	7.1
0.100	D	192.1	201.4	6.6
0.150	D	174.9	200.0	9.1
0.250	D	173.6	190.4	9.9
0.010	E	194.4	210.3	5.1
0.050	E	173.9	201.3	7.7
0.100	E	197.0	207.9	6.6
0.150	E	178.6	201.7	10.1
0.250	E	170.2	194.5	10.0

The recommended conditioning treatment for the RH 950 condition is to heat at 1750 $\pm$ 15 F for 10 minutes followed by air cooling. Figures 28 and 29 show schematically the effect of deviation from these conditions on the tensile properties of aged PH 15-7 Mo specimens. These show that the temperature can vary from about 1300 to 1800 F without appreciable effect, but that maximum strength is obtained within about 10 minutes at temperature. Longer heating periods result in increased ductility with some loss in strength. An investigation of the effect of conditioning temperature on fracture toughness indicated that higher precracked Charpy impact values at all test temperatures were obtained on specimens conditioned for 15 minutes at 1825 F, refrigerated at -90 F, and aged at 950 F. This is shown in Figure 30. Supplementary data showed that hardness and tensile strength were not affected by the same changes in austenite-conditioning temperature.

Refrigeration at -100 F is recommended to induce the transformation of austenite to martensite. Eight hours at -100 F is recommended and Figure 31 shows that the optimum RH 950 properties are obtained under such conditions. Refrigerating for longer periods, up to 24 hours, did not affect the tensile properties.

The effects of changes in the refrigeration cycle on the properties of 17-7 PH, RH 950 are shown in Figure 32. The graphs indicate that full properties may be developed under refrigerating conditions somewhat different from those in the recommended cycle. However, the latter are suggested to insure complete transformation if the M_s temperature happens to be lower than expected. The M_s temperature is quite sensitive to alloy composition and austenite conditioning sequences.

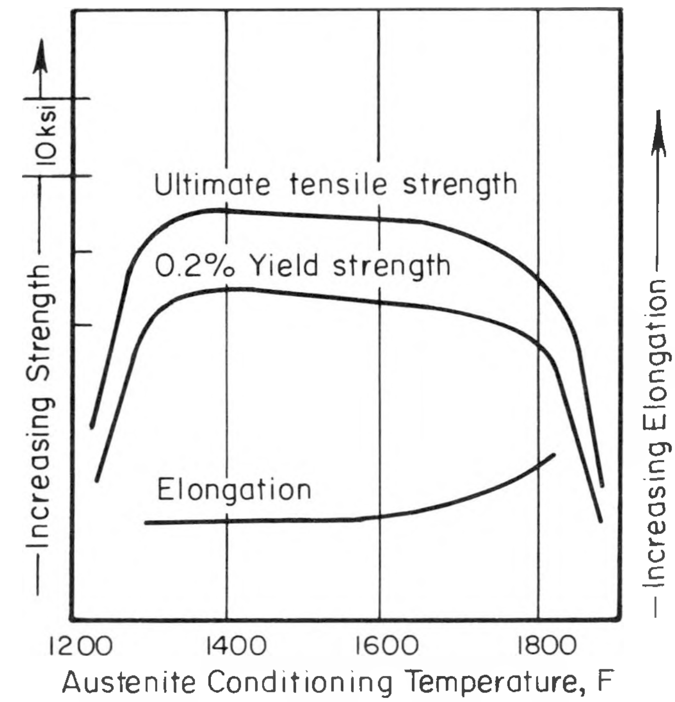


FIGURE 28. EFFECT OF AUSTENITE-CONDITIONING TEMPERATURE ON THE MECHANICAL PROPERTIES OF PH 15-7 Mo IN THE RH 950 CONDITION (Ref. 22)

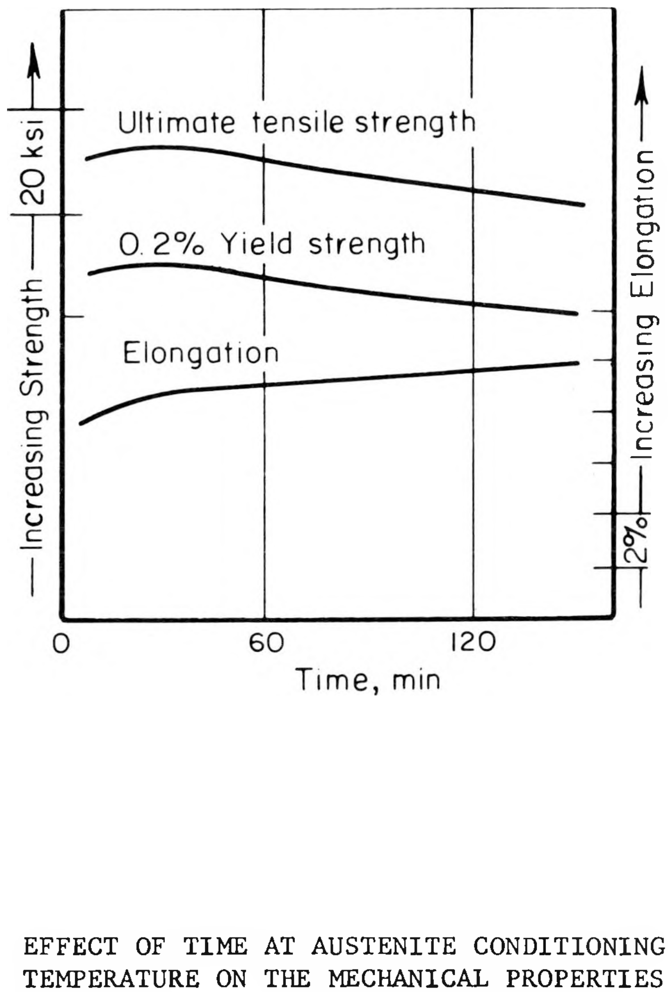


FIGURE 29. EFFECT OF TIME AT AUSTENITE CONDITIONING TEMPERATURE ON THE MECHANICAL PROPERTIES OF PH 15-7 Mo IN THE RH 950 CONDITION (Ref. 22)

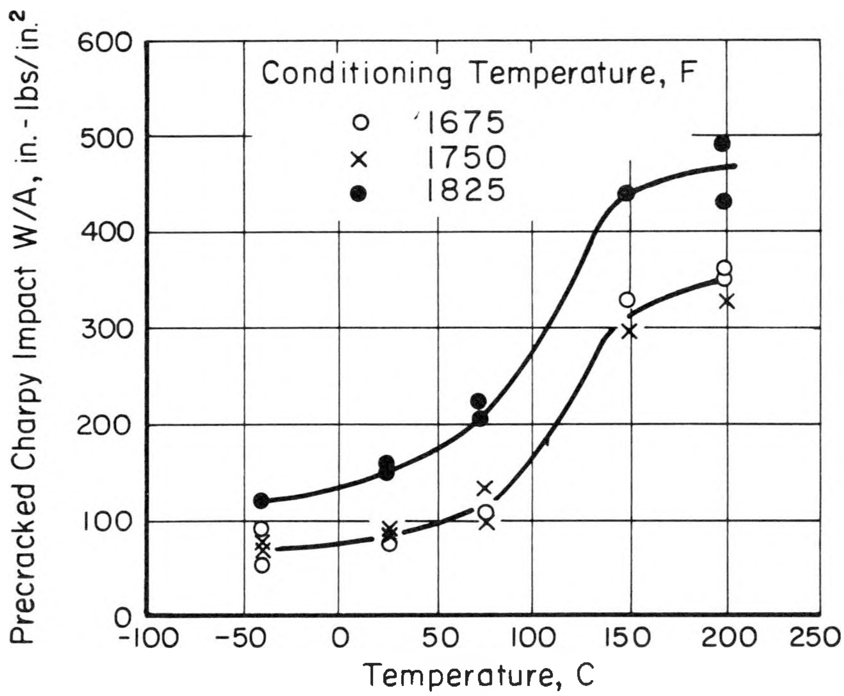


FIGURE 30. EFFECT OF CONDITIONING TEMPERATURE ON PRECRACKED CHARPY IMPACT TRANSITION TEMPERATURE CURVES FOR PH 15-7 Mo SHEET IN RH 950 CONDITION (Ref. 26)

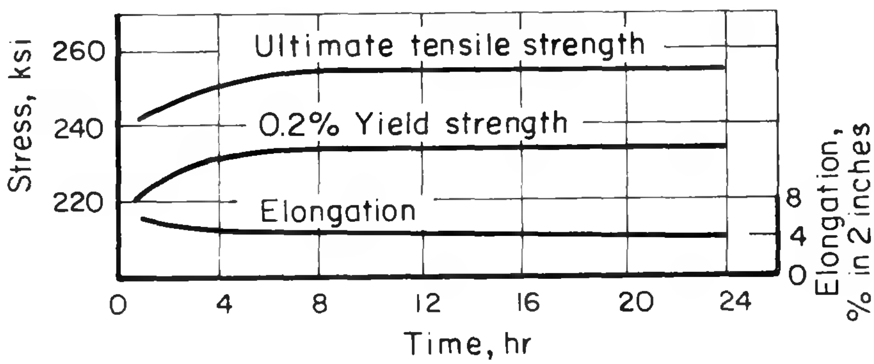


FIGURE 31. EFFECT OF VARIATION IN REFRIGERATION TIME AT -100 F ON MECHANICAL PROPERTIES OF PH 15-7 Mo, CONDITION RH 950 (Ref. 22)

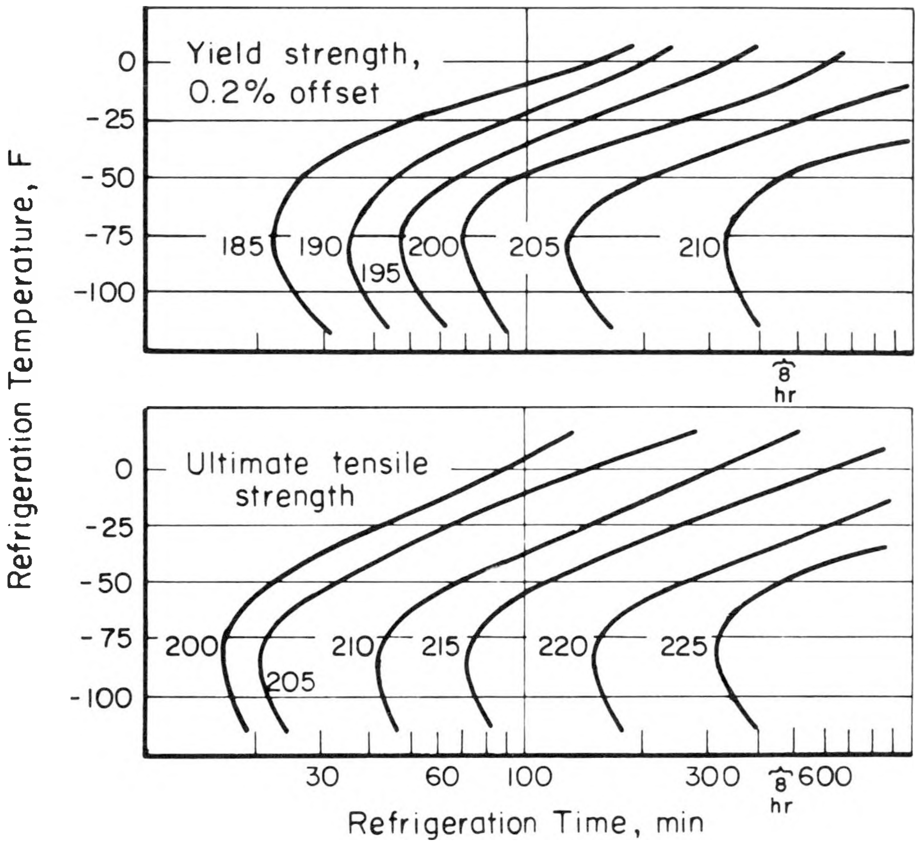


FIGURE 32. EFFECT OF REFRIGERATION CYCLE ON MECHANICAL PROPERTIES OF 17-7 PH, CONDITION RH 950 (Ref. 22)

(Held 10 minutes at 1750 F, air cooled to room temperature, liquid cooled to temperature shown, held for time shown, hardened at 950 F - 1 hour. Curves indicate strength in ksi)

Precipitation Hardening. The aging or precipitation hardening is the final step in the heat treatment sequence. The hardening effect takes place over a wide range of temperature, starting as low as 200 F and increasing to a maximum in the 900-950 F range. This is shown by the curves in Figure 33. Maximum strength, however, is accompanied by lowest ductility. The increased elongation and decreased strength of 17-7 PH resulting from raising the aging temperature above 950 F in the RH cycle treatment is shown in Figure 34. Similar trends at somewhat higher strength levels have been established for PH 15-7 Mo. The time at aging temperature has a similar effect on strength and ductility. Figures 35 and 36 show that maximum strength in 17-7 PH is developed within about 5 minutes at temperature, but elongation is at a minimum. Continued heating for 60 or 90 minutes results in somewhat lower strength and improved ductility. Figure 37 also shows the variation in tensile properties as a function of aging temperature in the RH treatment cycle. Maximum hardness and tensile properties are attained on aging at 950-1000 F. At these temperatures, the room-temperature fracture toughness drops to a minimum. On aging at temperatures above 950 F, the notched strength improves rapidly, but at increasing sacrifice of yield and tensile strengths (Ref. 27). Additional data on the effect of aging temperature on the notch strength are given in Table XXVIII, the $\frac{NS}{YS}$ and $\frac{NS}{UTS}$ ratios are increased. Relatively high aging temperatures should be used where good notch strength is required.

On occasion, it may be necessary or desirable to modify the properties obtained by using the standard TH 1050 or RH 950 treatments. This may be done by reaging at a different temperature from the initial

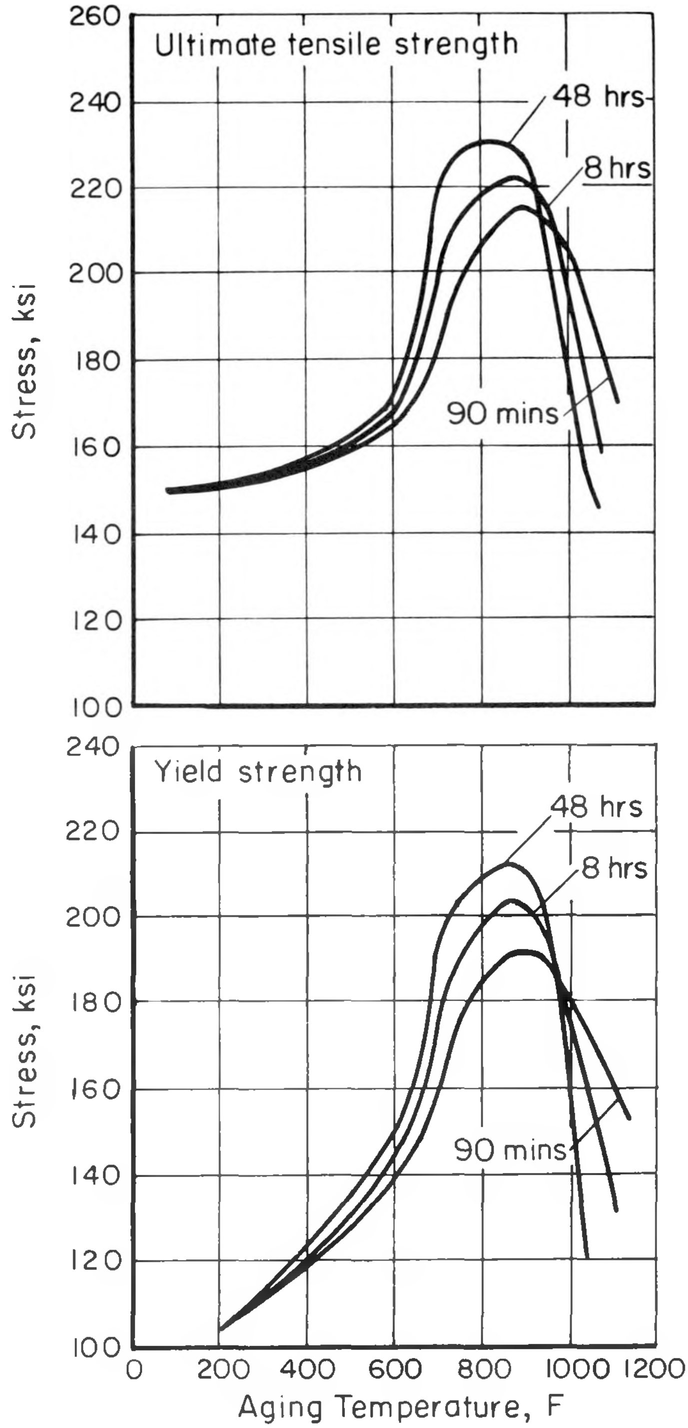


FIGURE 33. EFFECT OF AGING TIME AND TEMPERATURE ON MECHANICAL PROPERTIES OF 17-7 PH IN TH CONDITIONS (Ref. 22)

(1400 F, 1-1/2 hours, air cool to room temperature and water quench to 60 F. Age as shown.)

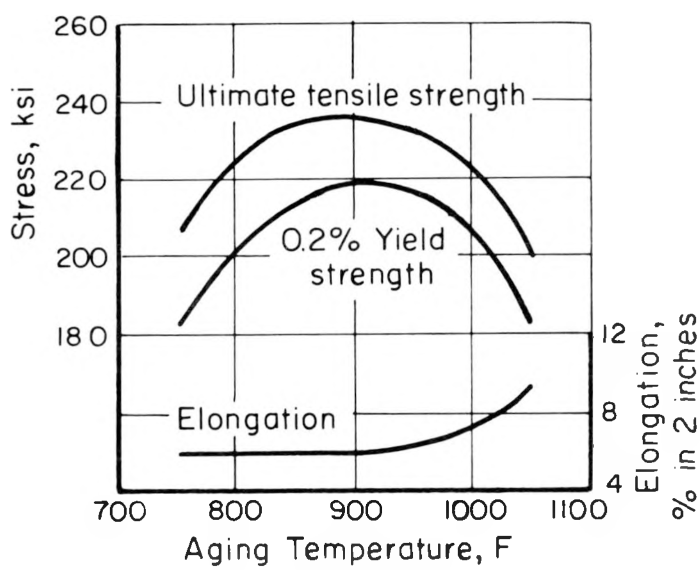


FIGURE 34. EFFECT OF AGING TEMPERATURE ON THE STRENGTH AND DUCTILITY OF 17-7 PH IN THE RH CONDITION (Ref. 22)

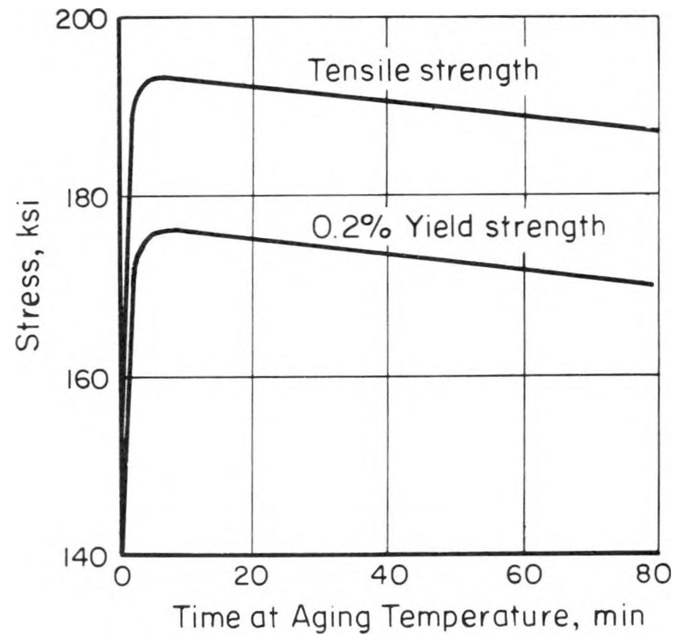


FIGURE 35. CHANGE IN STRENGTH OF 17-7 PH IN THE TH 1050 CONDITION WITH INCREASING TIME AT THE AGING TEMPERATURE (Ref. 22)

Tests in transverse direction

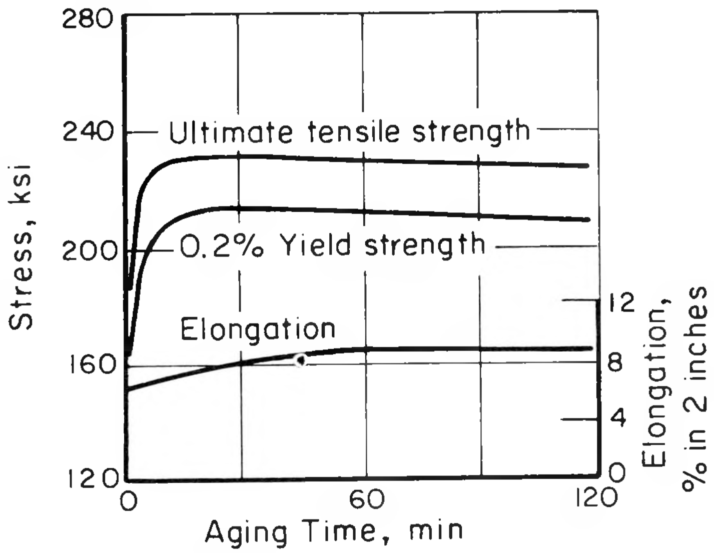


FIGURE 36. CHANGE IN TENSILE PROPERTIES OF 17-7 PH IN THE RH CONDITION WITH INCREASING TIME AT THE AGING TEMPERATURE (Ref. 22)

Data are average of two tests. Specimens 0.050 inch thick.

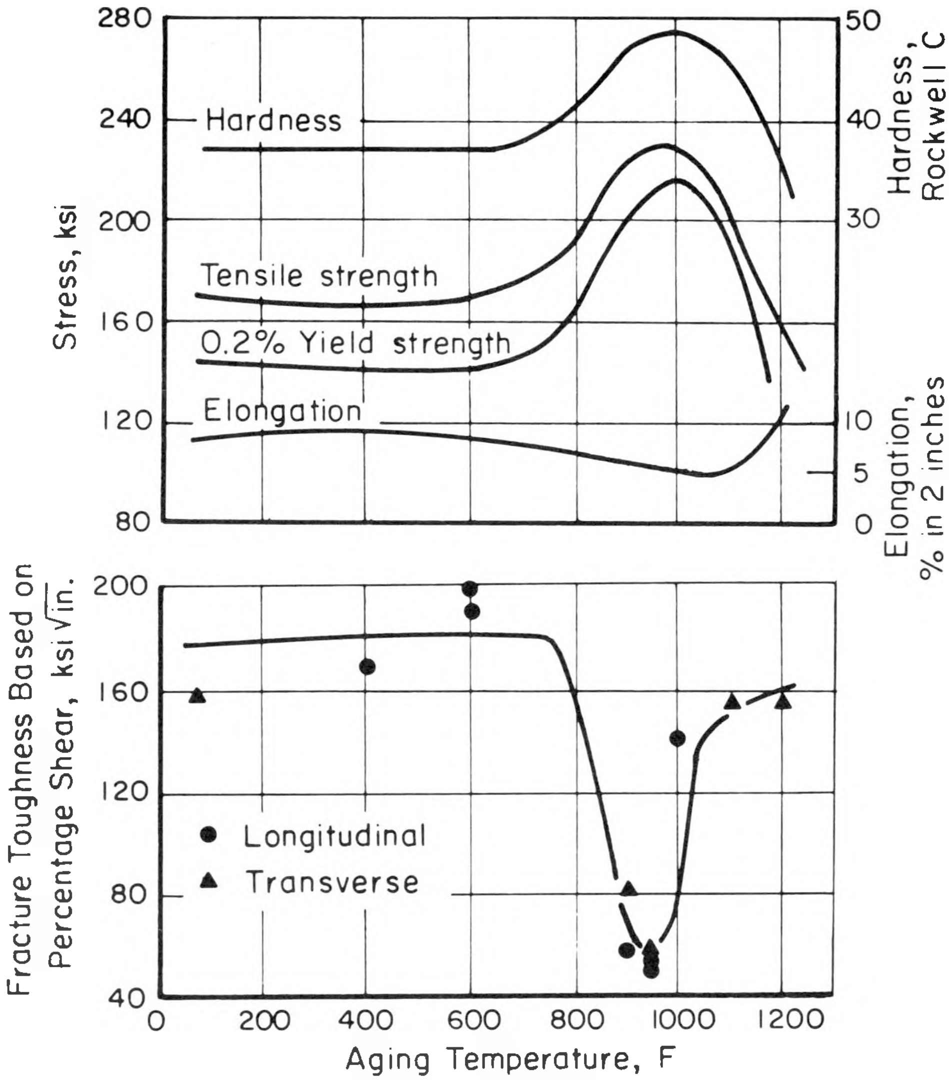


FIGURE 37. EFFECT OF AGING TEMPERATURE ON THE MECHANICAL PROPERTIES OF PH 15-7 Mo IN RH CONDITIONS (Ref. 27)

0.049-Inch-Thick Specimens Aged for 1 Hour

TABLE XXVIII. EFFECT OF AGING TEMPERATURE ON NOTCH STRENGTH OF PH 15-7 Mo (Ref. 22)

0.063-Inch Specimens, 1 Inch Wide, 0.0007-Inch Maximum Root Radius. Tests in Transverse Direction.

Precipitation Hardening Treatment	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Notch Strength, ksi	$\frac{YS}{NS}$	$\frac{UTS}{NS}$
RH 950	228	247	108	0.47	0.44
RH 1000	230	243	140	0.61	0.58
RH 1050	220	227	162	0.74	0.71
RH 1100	182	192	174	0.96	0.91

aging temperature. The effects of several combinations of rehardening treatments on the strength and ductility of PH 15-7 Mo in the TH and RH conditions are shown in Tables XXIX and XXX, respectively.

The practical application of such multiple aging treatments for 17-7 PH has been discussed by Token and Heldt (Ref. 24) in connection with the brazing of honeycomb panels. The heat of the brazing operation also serves to condition the austenite, but depending on the flow characteristics of the braze alloy, the temperature required for brazing may not be the one needed to develop adequate mechanical properties in the brazed assembly. In the example described by Token, the conditions of brazing and cooling apparently caused a shift in the M_s temperature which made it difficult to get the desired elongation after aging by the standard procedure. This was overcome by using a two-step aging procedure which improved the ductility while maintaining adequate strength in the panels. The full heat-treatment sequence, which includes provision for cooling and holding determined by practical shop procedures is as follows:

Braze: 1650 - 1690 F, 3 to 15 minutes
Cool to room temperature and hold for a minimum of 6 hours

Age: 1075 - 1100 F for 30 minutes, cool to 875 to 900 F and hold for 30 minutes

This treatment resulted in a tensile strength of 189.5 ksi, a yield strength of 177.8 ksi, and 7.1 percent elongation.

Cold Working. The properties of 17-7 PH in the 60 percent cold-worked and aged condition were given in Table XXV. These showed that the very high strength of the metal in Condition C was accompanied by low ductility which would limit or preclude forming. Additional data

TABLE XXIX. EFFECT OF AGING TEMPERATURE AND OF REAGING TREATMENTS ON THE MECHANICAL PROPERTIES OF PH 15-7 Mo, TH CONDITIONS (Ref. 22)

All Hardening Times 1-1/2 Hours Except As Noted

Precipitation-Hardening Treatment	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
TH 900 (30 minutes)	201	220	8.0	46.3
TH 950 (30 minutes)	209	224	7.5	47.0
TH 1000	220	231	6.7	48.2
TH 1050	205	216	6.2	46.5
TH 1050 + 1100	179	188	10.0	41.5
TH 1050 + 1050	201	206	6.7	44.8
TH 1050 + 1000	209	219	5.7	46.0
TH 1050 + 950	213	222	5.7	46.5
TH 1050 + 900	212	222	6.0	46.5
TH 1050 + 900 (65 hours)	223	234	5.2	48.0
TH 1100	181	189	8.0	41.5
TH 1100 + 1100	168	182	10.3	39.7
TH 1100 + 1050	182	191	9.3	41.7
TH 1100 + 1000	186	194	8.8	42.5
TH 1100 + 950	188	195	8.3	42.5
TH 1100 + 900	188	195	9.7	42.3
TH 1100 + 900 (65 hours)	202	210	8.5	44.7

TABLE XXX. EFFECTS OF AGING TEMPERATURE AND REAGING TREATMENTS ON THE MECHANICAL PROPERTIES OF PH 15-7 Mo IN THE RH CONDITIONS (Ref. 22)

Hardening Time 1 Hour Unless Noted Otherwise

Precipitation-Hardening Treatment	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
RH 900	211	237	7	48
RH 950	217	241	6	49
RH 950 + 1100	194	201	7	43
RH 950 + 1050	218	229	5	47
RH 950 + 1000	218	241	5	48
RH 950 + 900 (8 hours)	225	251	6	50
RH 950 + 900 (65 hours)	232	255	5	51
RH 950 + 900 (500 hours)	230	242	5	50
RH 1000	221	238	6	48
RH 1050	211	224	5	46
RH 1050 + 950	218	232	5	48
RH 1050 + 900 (8 hours)	224	236	5	49
RH 1050 + 900 (65 hours)	234	245	4	49
RH 1100	188	196	9	43

has been obtained to show the effect of intermediate amounts of cold working. The results are summarized in Table XXXI. It is clear that the properties are dependent on the degree of cold work. With 30 percent cold reduction (Condition 1/2 C) the alloy is still ductile enough to permit limited forming. Subsequent aging at 900 F results in somewhat lower strength than that obtained by aging the 60 percent cold-worked material. If the metal is cold worked after having been given the austenite-conditioning treatment, that is, in the A1750 condition, then refrigerated and aged to the RH 950 condition, the final properties obtained will depend on the amount of cold work. The more the cold work, the higher the strength and the lower the ductility. This is shown in Table XXXII. If material in the A 1750 condition is severely cold worked, it should be reconditioned at 1750 F to insure that adequate ductility will be attained after aging. If the cold working has been done on the metal in Condition A, followed by conditioning at 1750 F, there would have been no effect on the properties, because the austenite-conditioning step at 1750 F would have eliminated the martensite formed by cold working. In the case of material cold worked in Condition A followed by treatment to Condition TH 1050, it is found that the strength is decreased in proportion to the extent of cold working. This may be related to the stability and carbon content of the austenite existing after the 1400 F treatment. A modified TH 1050 treatment has been developed to overcome the loss of strength caused by cold work. The modified treatment is as follows:

Austenite Condition	1550 $\pm$ 25 F, 90 minutes, AC
Refrigerate	0 $\pm$ 10 F, 4 hours
Precipitation Harden	1050 $\pm$ 10 F, 90 minutes

TABLE XXXI. EFFECT OF VARIATION IN COLD REDUCTION ON ROOM-TEMPERATURE MECHANICAL PROPERTIES OF 17-7 PH (Ref. 22)

Cold Reduction, %	Thickness,(1) inch	Direction	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C	Bend Radius(2)
			As Rolled				
25	0.039	L	112.6	167.0	14.8	35	0.75 T
		T	104.2	170.0	12.5		2.25 T
30	0.0365	L	120.0	173.0	14	36.5	1.25 T
		T	114.7	178.0	10		3.5 T
35	0.034	L	137.2	174.6	12	38	1.5 T
		T	118.9	183.7	9.2		5 T
40	0.031	L	160.0	186.5	10	39	2.5 T
		T	133.9	197.2	8		6.5 T
45	0.029	L	164.0	187.4	10	40	2.5 T
		T	148.2	201.2	7.8		8 T
			Aged 900 F - 1 Hour				
25	0.039	L	162.8	169.8	17.8	41	1.25 T
		T	164.6	179.0	15		3 T
30	0.0365	L	174.2	177.2	11	43	1.25 T
		T	176.3	191.1	13.5		4 T
35	0.034	L	191.1	196.6	6.2	44.5	1.75 T
		T	191.0	200.1	12		4.5 T
40	0.031	L	222.5	229.0	4.0	47	2.5 T
		T	218.2	225.5	6.5		6.5 T
45	0.029	L	225.0	227.3	2.0	48	2.5 T
		T	229.0	233.4	4.5		8 T

(1) Initial thickness 0.0525 inch. Data as based on duplicate tests from one heat.

(2) Radius for 180 degrees free bend without cracking.

TABLE XXXII. EFFECT OF COLD WORK ON 17-7 PH, CONDITION RH 950 PROPERTIES (Ref. 22)

17-7 PH Samples 0.050-Inch Thick Were Stretched While in Condition A 1750, Then Refrigerated at -100 F for 8 Hours and Aged at 950 F 1 Hour

Stretch, %	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
0	214	226	8	47
3.5	219	228	6	47
6.0	220	229	6	46
8.0	216	237	5	47
10.0	241	246	3	48
11.0	241	244	4	48

Harrington (Ref. 13) has reported that significant increase in strength and ductility may be obtained by "stress aging" 17-7 PH TH 1050 at temperatures of 550 to 750 F for periods of 1/4 to 4 hours. The alloy is heated while under applied tensile stresses approaching its 0.1 percent offset yield strength in the TH 1050 condition. The change in properties caused by one such "stress-aging" treatment is shown in Figure 38. The treatment resulted in an increase in tensile and yield strength as well as ductility, with the greatest increase being obtained in the proportional limit.

The effect of a change in aging temperature on the hardness of PH 15-7 Mo, initially cold-rolled 60 percent, is shown in Figure 39. This shows that an increase in aging temperature above the recommended value of 900 F causes a decrease in the hardness of the alloy. This figure also provides a comparison of the hardness obtained in the RH and CH conditions.

Material cold worked after solution heat treatment (Condition A) and then conditioned and hardened to Condition TH 1050 does

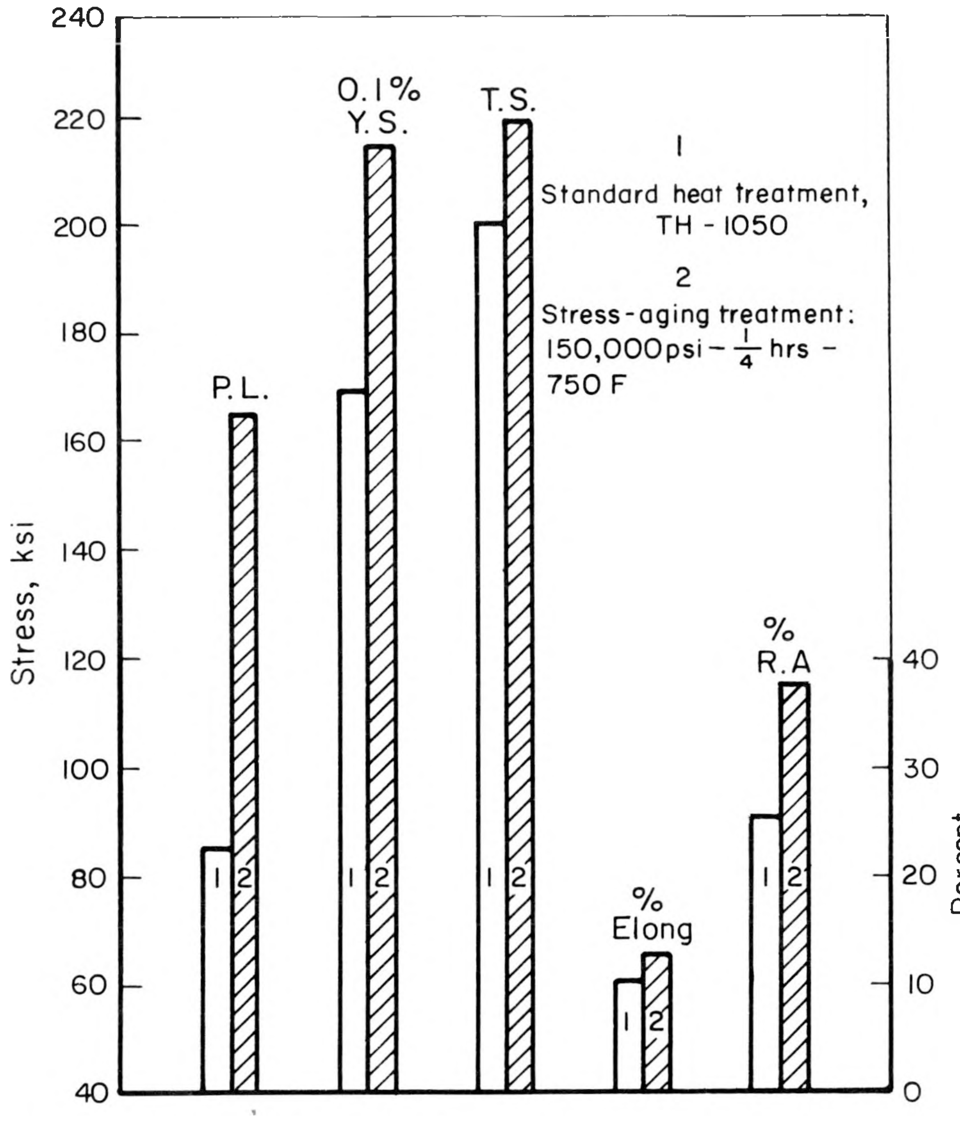


FIGURE 38. EFFECT ON TENSILE PROPERTIES OF A STRESS-AGING TREATMENT APPLIED TO 17-7 PH, TH 1050 BAR STOCK (Ref. 13)

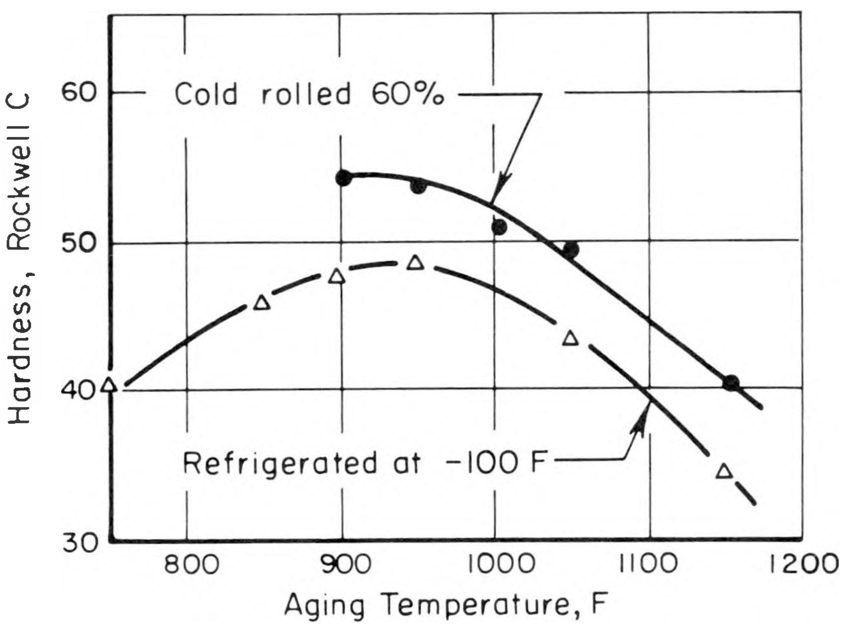


FIGURE 39. HARDNESS OF PH 15-7 Mo AFTER AGING ONE HOUR AT VARIOUS TEMPERATURES (Ref. 27)

0.049-Inch Sheet Initially Cold Rolled 60% or Refrigerated not attain the optimum strength properties, but instead decreases in strength according to the amount of cold work. This is thought to be related to the greater stability and higher carbon content of the austenite existing after the 1400 F conditioning treatment. The following tabulation shows the relationship between degree of cold work and the properties obtained following the TH 1050 treatment.

Reduction in Thickness by Rolling, %	Yield Strength, 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
0	188	199	9.5	44.5
20	163	191	10.8	43.0
30	150	189	11.4	42.5

A modified treatment has been developed to prevent the loss in strength caused by the cold working. This is as follows:

Austenite Conditioning	1550 $\pm$ 25 F, 90 minutes, AC
Refrigerate	0 $\pm$ 10 F, 4 hours
Precipitation Harden	1050 $\pm$ 10 F, 90 minutes

The influence of various amounts of cold work, and cold working followed by aging on mechanical properties, has been studied by Armco. The data obtained are assembled in Table XXXIII. The effects on the yield strength are plotted in Figure 40 . It is quite clear that the properties obtained are determined by the type of thermal treatments that follow the cold-rolling operation.

A comparison of the fracture toughness properties of PH 15-7 Mo in the RH and CH conditions over a range of temperatures is given in Figure 41. The aging treatments for the cold-rolled material were adjusted to give the same room-temperature yield strengths as obtained by the RH 950 in both the longitudinal and transverse directions. The results show that the room-temperature notch toughness of the longitudinal cold-worked and aged material is much higher than that of the RH 950 material (227,000 psi $\sqrt{in.}$ vs. 51,000 psi $\sqrt{in.}$). Furthermore, the cold-worked material

TABLE XXXIII. EFFECT OF COLD WORK ON MECHANICAL PROPERTIES OF PH 15-7 Mo (Ref. 22)

Cold Rolling Temperature F	Condition Tested	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
0% CR	A	53	127	38.5	85.5 B
	900 1 hour	46	128	44.0	89.5 B
	LH 950[a]	213	233	6.5	49.0
		20% Cold Reduction			
32	CR	122	175	12.0	38.0
	CR + 900	166	175	19.0	43.0
	CR + LH 950	205	223	7.0	49.5
100	CR	108	162	16.5	35.5
	CR + 900	141	149	18.0	38.5
	CR + LH 950	220	235	5.0	49.5
200	CR	101	154	19.0	32.5
	CR + 900	131	138	20.0	36.0
	CR + LH 950	217	233	4.0	50.5
300	CR	115	148	22.0	31.5
	CR + 900	128	137	20.0	35.0
	CR + LH 950	226	242	5.0	50.5
		40% Cold Reduction			
32	CR	194	198	3.0	45.0
	CR + 900	258	265	2.0	51.5
	CR + LH 950	176	199	11.0	46.0
100	CR	154	184	10.0	42.5
	CR + 900	210	228	3.0	49.0
	CR + LH 950	201	221	5.5	49.0
200	CR	124	171	16.0	38.5
	CR + 900	182	185	4.5	45.0
	CR + LH 950	221	237	4.5	50.5
300	CR	104	162	16.5	34.5
	CR + 900	146	150	16.5	39.5
	CR + LH 950	230	242	5.0	50.5

TABLE XXXIII. Continued

Cold Rolling Temperature F	Condition Tested	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
		60% Cold Reduction			
32(b)	CR	221	225	1.0	46.0
	CR + 900	288	292	2.0	54.0
	CR + LH 950	176	191	14.0	46.0
100	CR	221	225	2.0	46.0
	CR + 900	286	288	2.0	52.5
	CR + LH 950	181	197	7.0	46.5
200	CR	166	188	10.0	43.0
	CR + 900	231	234	3.5	48.5
	CR + LH 950	205	222	4.0	49.0
300	CR	127	173	16.0	40.5
	CR + 900	181	187	5.5	44.5
	CR + LH 950	234	246	3.0	51.0
		76% Cold Reduction			
100	CR	244	252	1.5	48.5
	CR + 900	313	319	1.5	55.5
	CR + LH 950	171	186	15.0	46.0
200	CR	217	233	3.0	48.0
	CR + 900	286	296	1.0	54.0
	CR + LH 950	178	194	9.5	46.5
300	CR	154	190	13.5	44.0
	CR + 900	213	225	4.5	48.5
	CR + LH 950	226	244	4.0	51.0

(a) LH 950 treatment: 1200 F/2 hours, -100 F/8 hours, 950 F/1 hour

(b) 50% cold reduction

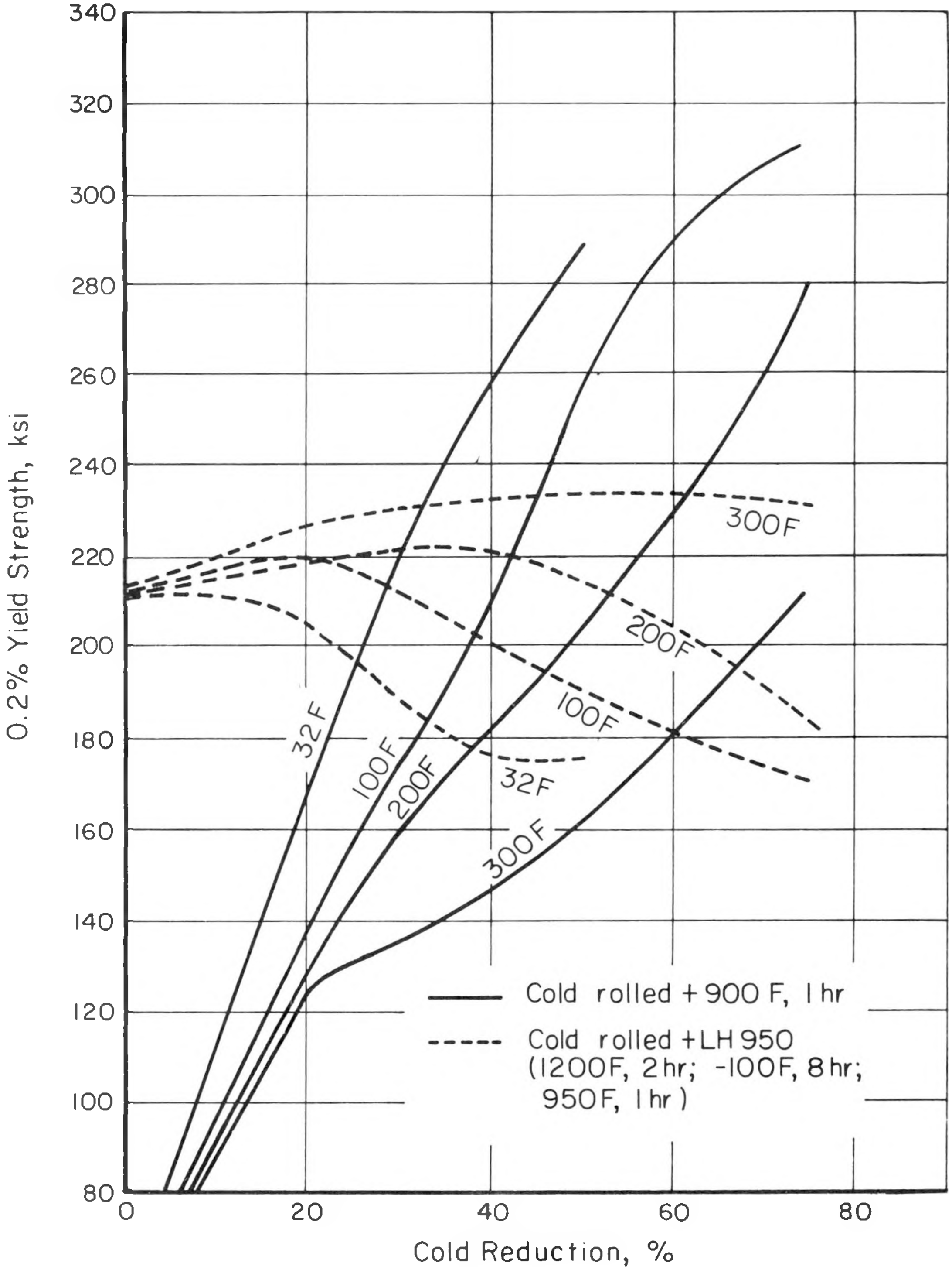


FIGURE 40. EFFECT OF PERCENT COLD REDUCTION AND COLD REDUCTION TEMPERATURE ON YIELD STRENGTH OF PH 15-7 Mo (Ref. 22)

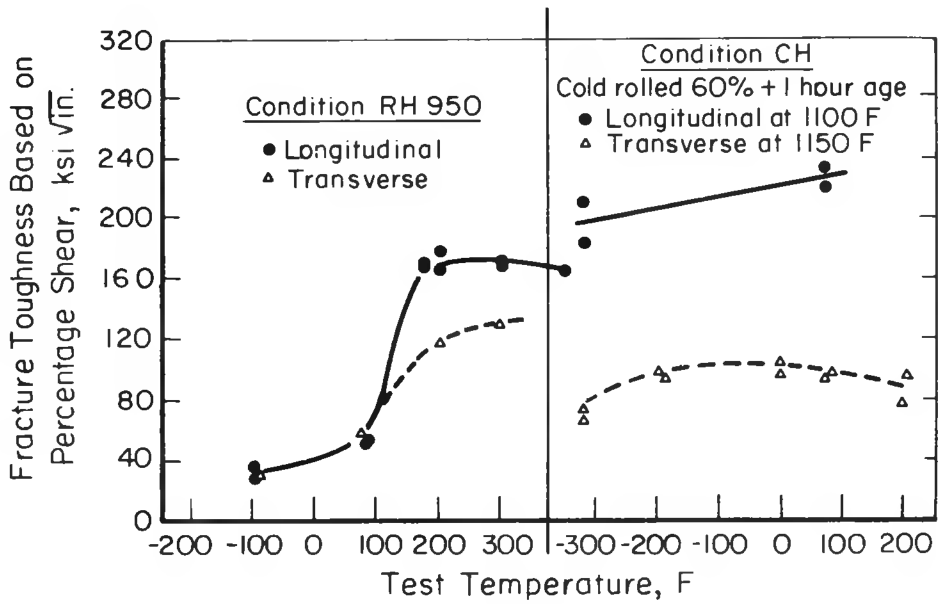


FIGURE 41. EFFECT OF THERMAL AND MECHANICAL TREATMENTS ON THE FRACTURE TOUGHNESS OF PH 15-7 Mo (Ref. 27)

Specimens 0.049 inch thick. Aging treatments for the cold-rolled material adjusted as indicated to provide equal room-temperature yield strengths.

retains good notch toughness at low temperatures to -320 F, whereas the toughness of the RH material decreases rapidly with temperature. The differences in the transverse direction are not so great.

PH 14-8 Mo. This is a relatively new semiaustenitic PH stainless steel that can be treated to provide high strength coupled with fracture toughness (Ref. 28). The chemical composition of PH 14-8 Mo is similar to that of 15-7 Mo (see Table I). The major differences in PH 14-8 Mo are lower carbon, phosphorus and sulfur contents, and a slightly lower chromium and slightly higher nickel content. The resulting metallurgical structure is similar to that of the other semiaustenitic PH steels, but the grain boundaries are relatively free of carbides, which is an important factor in promoting improved resistance to crack propagation. The improved properties are obtained with air-melted material, but an additional increase in toughness is obtained with vacuum-melted material, brought about by accurate control of chemical composition (Ref. 29).

The annealed and fabricated material may be conditioned, transformed and precipitation hardened according to the standard procedures illustrated in Figure 42. Typical properties developed by these treatments for air-melted and vacuum-induction-melted steel are listed in Table XXXIV. These are the properties to be expected for each condition when using the recommended procedures. Intentional or unintentional deviation from the standard treatments will have an effect on the properties obtainable in the fully heat-treated steel. The available data on such effects are given in the succeeding paragraphs.

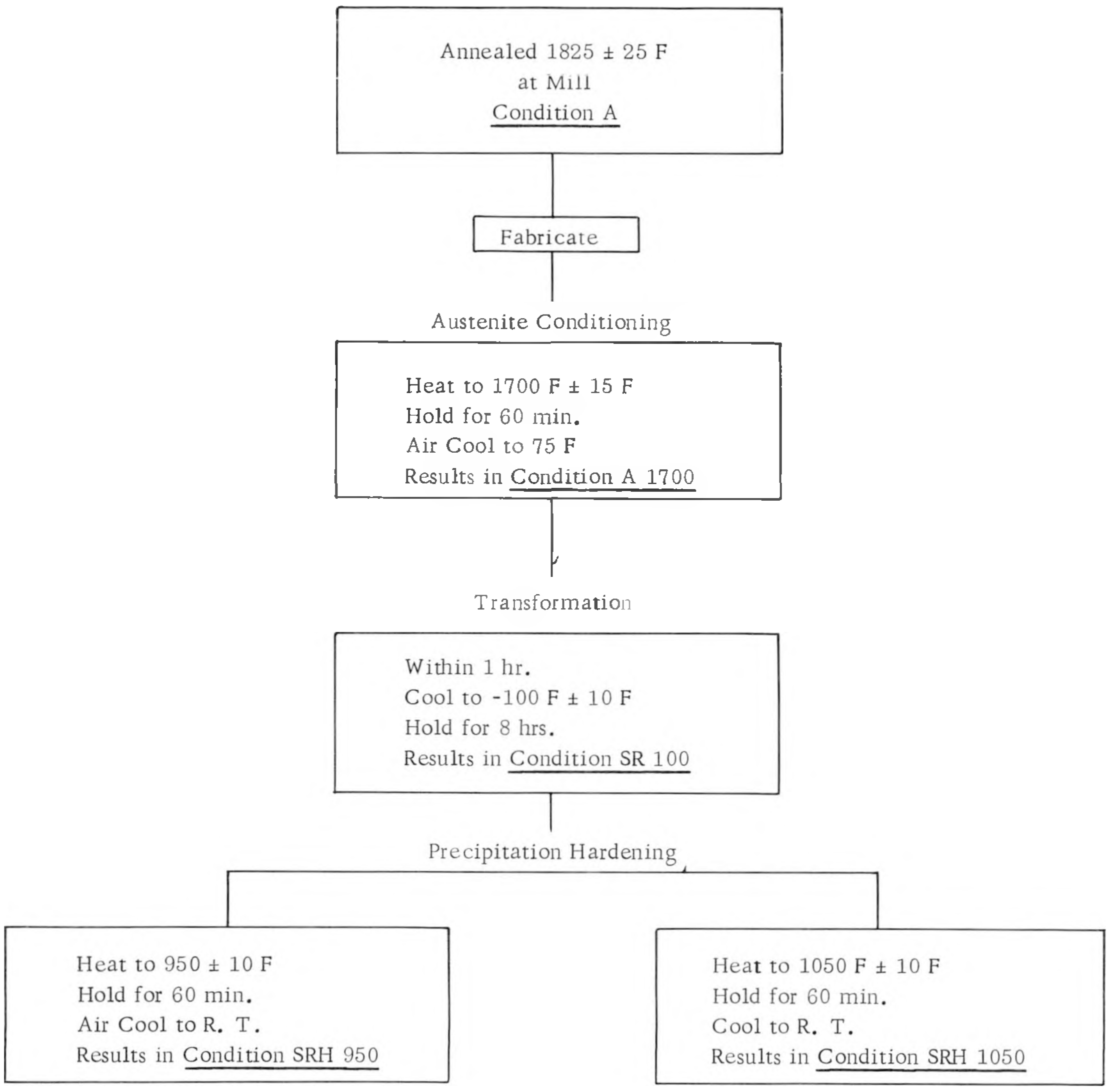


FIGURE 42. STANDARD HEAT TREATMENTS FOR PH 14-8 Mo (Ref. 28)

TABLE XXXIV. TYPICAL ROOM-TEMPERATURE MECHANICAL PROPERTIES OF PH 14-8 Mo OBTAINED BY STANDARD HEAT TREATMENTS (Ref. 28)

Tested in Transverse Direction

Condition	Yield Strength, 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell	Toughness,[a] W/A in-lbs/in^2
Air Melted					
A	55.0	125.0	25	B 88	-
SRH 950	220.0	235.0	5	C 49	850
SRH 1050	205.0	215.0	5	C 46	1000
Vacuum Induction Melted					
A	55.0	125.0	25	B 88	-
SRH 950	215.0	230.0	6	C 48	1800
SRH 1050	200.0	210.0	6	C 45	2200

(a) Precracked Sheet Charpy Specimen: Same as ASTM E 23-64, Figure 4A, except thickness of specimen is sheet thickness and specimen is precracked by fatigue to a depth of 0.030-0.060 inch.

Variation in Annealing. The recommended annealing temperature is 1825 ± 25 F. Figure 43 shows the influence of changes in annealing temperature between 1700 and 2150 F on the tensile and yield strengths, both as annealed (Condition A) and after a subsequent age-hardening treatment (Condition SRH 950). This indicates that the annealing temperature can vary over a wide range without affecting the strength of the steel aged to the SRH 950 condition.

Variation in Conditioning Treatment. The recommended austenite conditioning temperature is 1700 F, but it has been reported (Ref. 30) that variation over the range from 1450 to 1750 F has little effect on the strength level of the precipitation-hardened material. However, the time at temperature is important. Figure 44 shows the spread in strength that is likely to be obtained for any given time at temperature. Evidently the variation in strength is very large for short conditioning times (20 to 40 minutes). The results are more reproducible with a conditioning time of 60 minutes. Conditioning temperature does affect the toughness, however, as can be seen in Figure 45. Therefore, the optimum level of strength and toughness is obtained when a conditioning treatment of 1700 F for one hour is used.

The effect of conditioning temperature on the room and cryogenic temperature toughness of PH 14-8 Mo SRH 1050 is shown by data in Table XXXV.

Cooling rate from the austenite-conditioning temperature to 1000 F is another important variable. Figure 46 indicates that slow cooling from 1700 F causes a marked reduction in strength of some heats in the SRH 1050 condition. This apparently does not occur if the austenite-

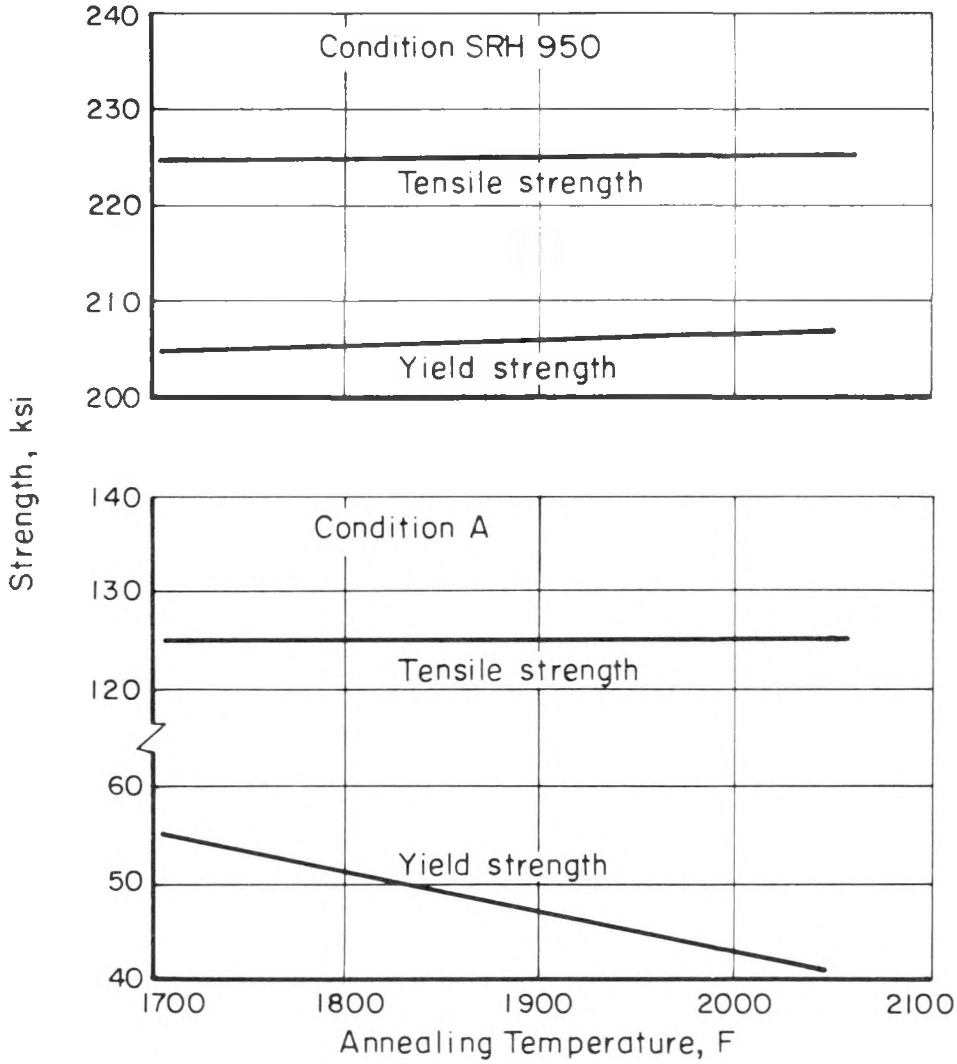


FIGURE 43. EFFECT OF VARIATION IN ANNEALING ON STRENGTH OF PH 14-8 Mo (Ref. 30)

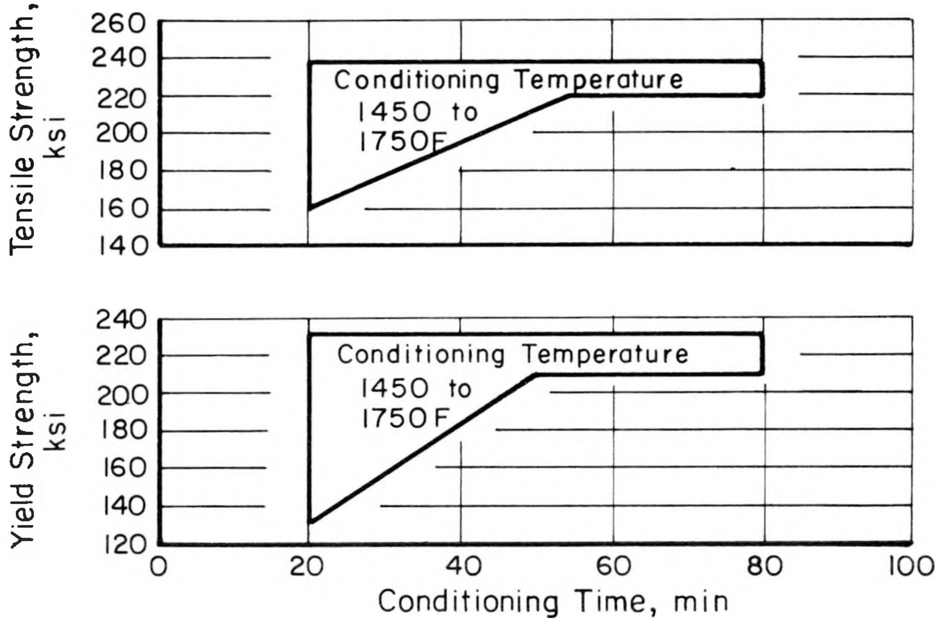


FIGURE 44. INFLUENCE OF CHANGES IN CONDITIONING TIME AND TEMPERATURE ON THE TENSILE AND YIELD STRENGTH OF PH 14-8 Mo, SRH 950 (Ref. 30)

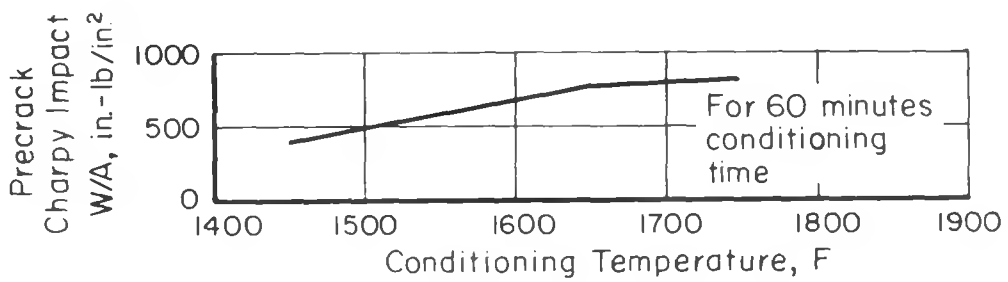


FIGURE 45. EFFECT OF AUSTENITE CONDITIONING TEMPERATURE ON THE TOUGHNESS OF PH 14-8 Mo SRH 950 (Ref. 30)

TABLE XXXV. EFFECT OF CONDITIONING TEMPERATURE ON ROOM AND CRYOGENIC TOUGHNESS OF PH 14-8 Mo (Ref. 30)

Transverse direction: Sheet 0.040 to 0.060 inch thick; SRH 1050 condition

	Precracked Sheet Charpy	
Conditioning Temperature	W/A $in\text{-}lb/in^2$ at 75 F	W/A $in\text{-}lb/in^2$ at -300 F
1600 F	908	409
1650 F	1077	622
1700 F	1334	877

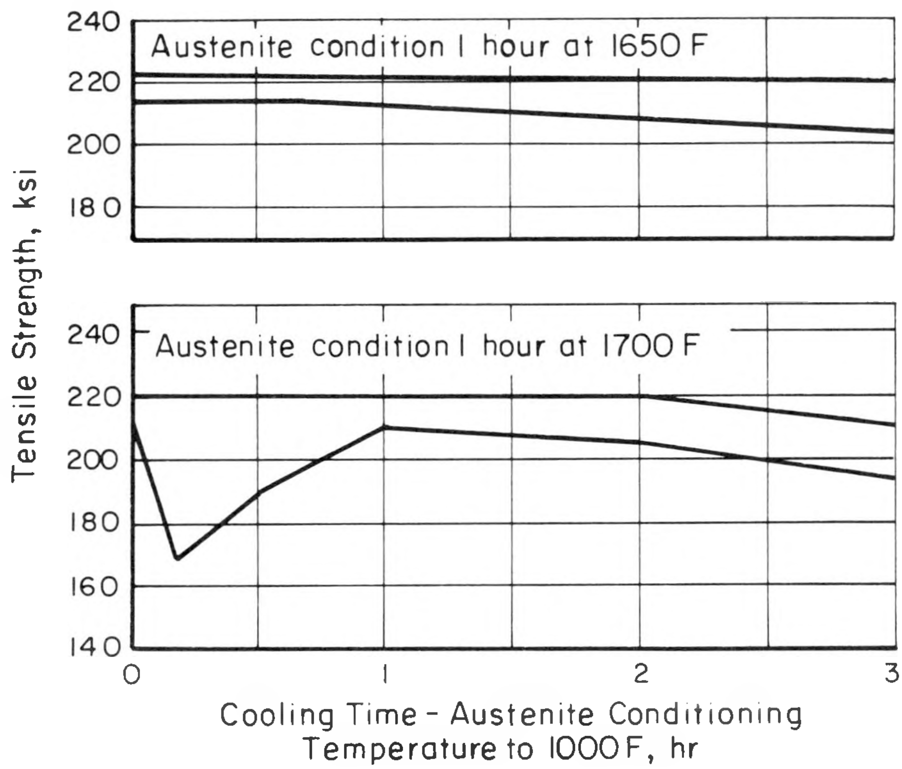


FIGURE 46. EFFECT OF VARIATION IN COOLING TIME AFTER AUSTENITE CONDITIONING ON THE TENSILE STRENGTH OF PH 14-8 Mo, SRH 1050 (Ref. 30)

conditioning temperature is reduced only 50 degrees to 1650 F. However, the cooling time, even after conditioning at 1650 F, should be as short as possible to insure maximum toughness. The spread in precracked sheet Charpy values obtained as a function of cooling rate is shown in Figure 47.

Variation in Transformation Conditions. The standard recommended conditions to effect transformation of austenite to martensite are to refrigerate at -100 F for 8 hours. The effect of refrigerating at lower or higher temperature and for different lengths of time is illustrated in Figure 48. At -50, -100 and -150 F the strength increases slightly in proportion to the refrigeration time. At -200 F, adequate transformation apparently does not take place in less than 8 hours. The optimum temperature appears to be -100 F as recommended.

Variation in Aging Treatment. Figure 49 shows the effect of a few variations in aging temperature and time on the strength and toughness of PH 14-8 Mo. Long aging time (7 hours) at temperatures above about 925 F results in slightly lower tensile and yield strengths than obtained by the standard 1-hour aging. Toughness is improved by both longer aging, and higher aging temperature.

Cold Working. In common with 17-7 PH and PH 15-7 Mo, cold working also may be used on PH 14-8 Mo to obtain very high strength. Heavy cold reduction at the mill, and cold reduction followed by aging at 900 F for 1 hour are designated as Condition C and Condition CH 900, respectively. The effect of the amount of cold reduction, as cold rolled and after aging one hour at 900 F, on the tensile properties is shown in Table XXXVI. Tensile strengths of about 200,000 psi may be obtained by

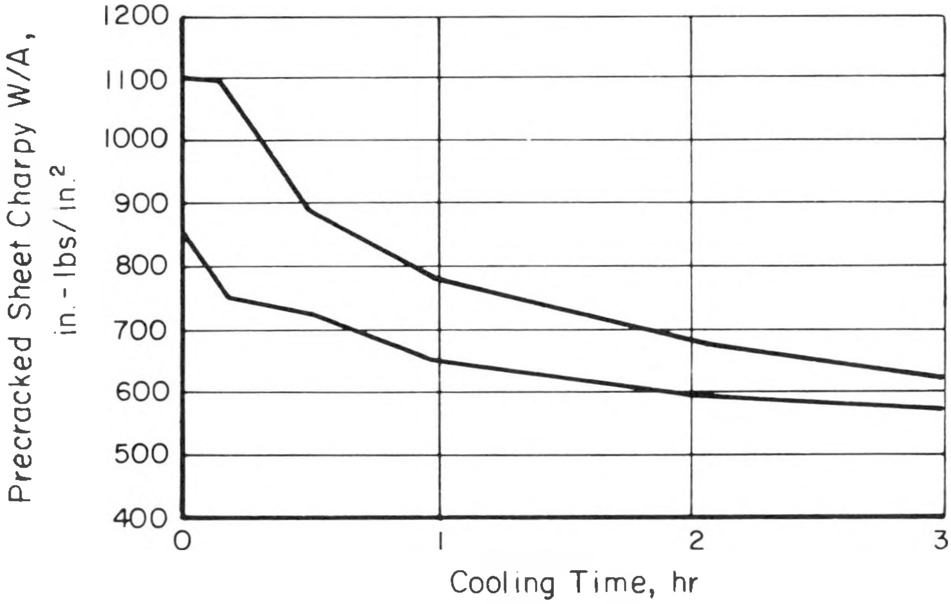


FIGURE 47. EFFECT OF COOLING RATE AFTER AUSTENITE CONDITIONING ON TOUGHNESS OF PH 14-8 Mo, SRH 1050 (Ref. 30)

Air-melted sheet; conditioning temperature 1650 F. Curves delineate the spread in values.

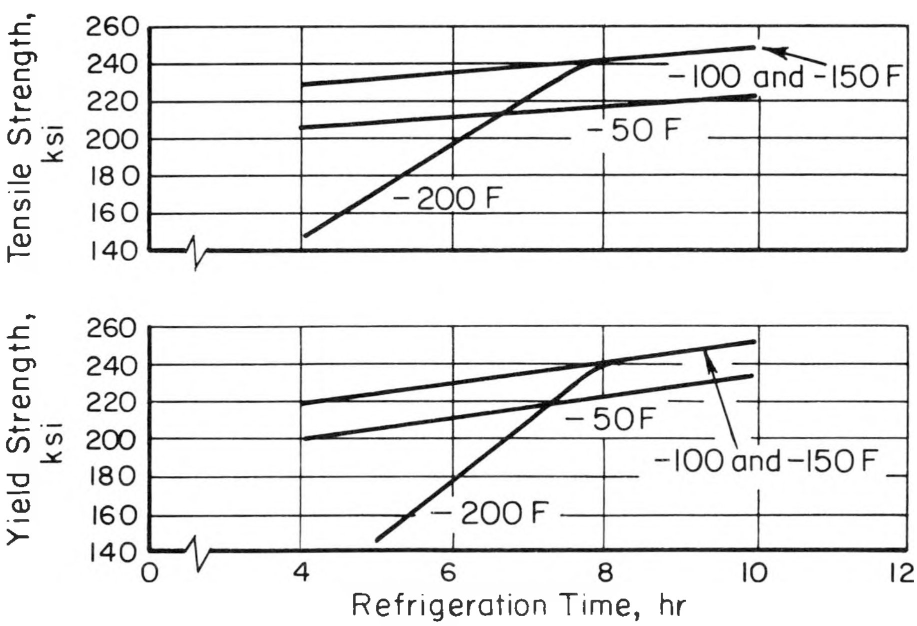


FIGURE 48. EFFECT OF REFRIGERATION TEMPERATURE AND TIME ON THE TENSILE STRENGTH OF PH 14-8 Mo, SRH 950

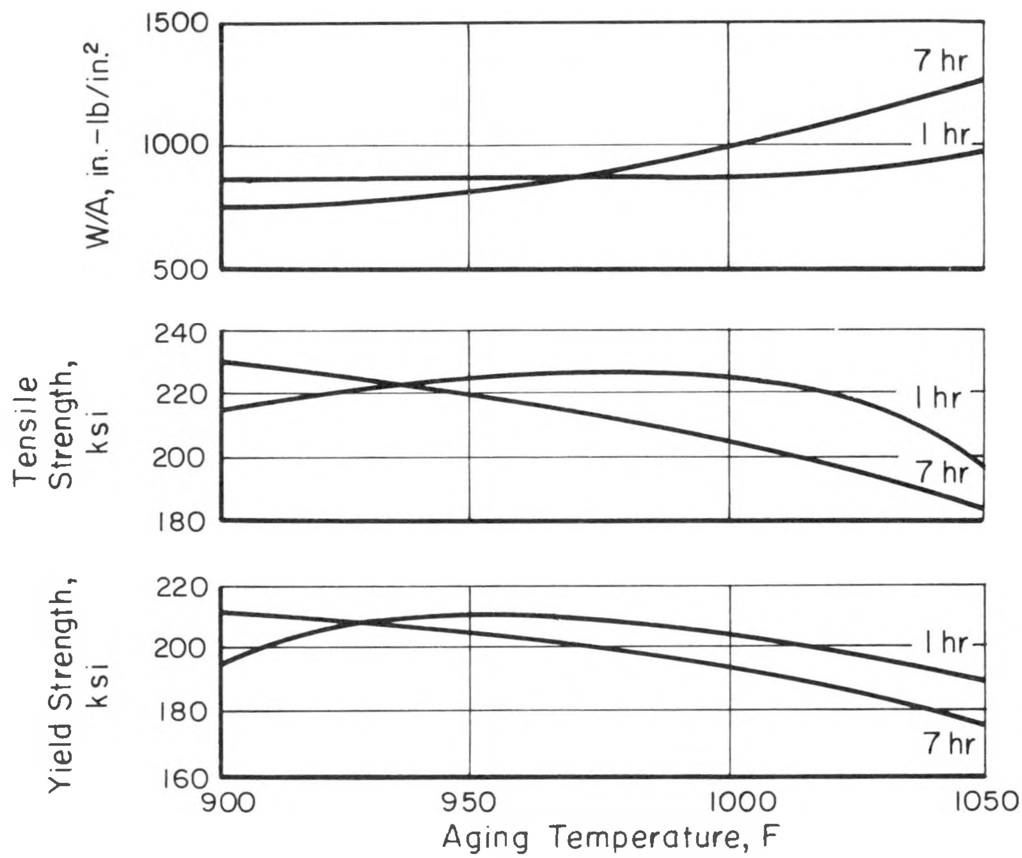


FIGURE 49. MECHANICAL PROPERTIES OF PH 14-8 Mo VS. AGING TEMPERATURE AND TIME (Ref. 30)

Standard conditioning and transformation treatments

TABLE XXXVI. EFFECT OF COLD REDUCTION ON THE TENSILE PROPERTIES OF PH 14-8 Mo IN CONDITION C AND CONDITION CH 900 (Ref. 28)

Cold Reduction, %	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
		As Cold Rolled		
6.3	61	131	23.0	90 B
11.3	65	141	16.5	21.5
22.2	78	163	11.0	33.0
29.2	109	168	10.0	38.0
39.2	150	178	8.0	41.0
49.9	177	193	2.0	43.0
59.7	207	209	1.0	43.5
72.5	222	225	1.0	45.0
		Cold Rolled + 900 F, 1 Hour		
6.3	71	129	28.5	88.5 B
11.3	94	130	23.5	24.5
22.2	135	158	15.5	39.5
29.2	206	208	5.0	47.5
39.2	250	252	2.0	50.5
49.9	276	277	1.0	53.0
59.7	290	290	1.0	54.0
72.5	305	310	OG*	55.5

* Specimens broke outside of gage

cold rolling alone, and these values can be raised to 300,000 psi by aging at 900 F. As might be expected, the ductility under these conditions is very low. Additional data, showing the effect of cold reduction on longitudinal and transverse tensile properties after aging at 950 F (Condition CH 950), compared also with the alloy hardened by transformation and aging at 950 F (Condition SRH 950) are given in Table XXXVII. The improved room-temperature notch strength of the cold-worked material, particularly in the transverse direction, brought about by vacuum-induction melting is evident from the data in Table XXXVII.

Cold working following an austenite-conditioning treatment promotes even greater strengthening than cold rolling the annealed material. This is shown by the data in Table XXXVIII, compared with the data in XXXVI. The austenite-conditioning treatment used in this investigation was not the normal 1-hour heating at 1700 F, but was intended to simulate the temperature conditions existing on strip moving through a furnace continuously at fairly high speed. The metal was at temperature for only about 30 seconds, and this accounts for the relatively minor effect on the properties noted on specimens that were not cold worked, or subjected to only slight working after conditioning.

Braze-Cycle Treatments. Numerous braze-cycle heat treatments have been developed or tested. These usually incorporate heating and cooling treatments which are compatible with the requirements for satisfactory brazing and for transformation and age hardening of the alloy to develop suitable mechanical properties. The melting characteristics of the brazing alloy determine the temperature that must be used for brazing. This temperature may be in the solution treating or austenite-conditioning range for the semiaustenitic stainless steels. Therefore, the brazing

TABLE XXXVII. EFFECT OF COLD REDUCTION ON THE TENSILE PROPERTIES OF PH 14-8 Mo (Ref. 30)

Cold Reduction, %	Direction	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C	Notch Strength, ksi	Notch Strength / Yield Strength	Notch Strength / Ultimate Tensile Strength
Air Melted, SRH 950								
0	L	212	232	8	-	214	1.01	0.92
	T	217	235	5	-	193	0.89	0.82
Air Melted, CH 950								
40	L	254	259	2.0	51	253	1.00	0.98
	T	260	270	1.5	51	206	0.79	0.76
60	L	273	278	2.0	52	237	0.87	0.85
	T	289	298	1.5	52	154	0.53	0.52
Vacuum-Induction Melted, CH 950								
40	L	235	240	1.5	50	241	1.02	1.00
	T	256	261	1.5	50	241	0.94	0.92
60	L	258	264	1.5	50	249	0.96	0.94
	T	269	279	1.5	50	237	0.88	0.84

NASA 1-inch-wide edge-notch specimen.
0.0007-inch maximum root radius on 0.025-inch-thick sheet.

TABLE XXXVII. EFFECT OF COLD WORK AFTER AUSTENITE CONDITIONING ON THE TENSILE PROPERTIES OF PH 14-8 Mo (Ref. 30)

Conditioning[a]	Cold Reduction %	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in, %	Hardness, Rockwell C
		As Cold Rolled			
1700 F	0	81.0	147.2	20.5	24
	10.7	85.1	158.2	15	32
	17.2	125.8	174.1	10	38
	27.9	161.7	182.2	6.5	
1700 F + -50 F 3 hours	0	72.7	140.6	24	21
	9.9	94.0	161.9	13	33
	16.5	142.8	174.7	8	39
	27.3	179.5	194.7	4	40
		Cold Rolled + 2 Hours at 950 F			
1700 F	0	89.7	147.9	23	26
	10.7	156.0	169.4	15.5	40
	17.2	217.5	232.2	3.5	48
	27.9	249.1	258.9	3	50
1700 F + -50 F 3 hours	0	79.8	144.0	25	22
	9.9	149.6	177.6	7	42
	16.5	222.0	241.7	3	49
	27.3	264.0	269.7	3	52

(a) Simulated strip conditioning -metal at conditioning temperature about 30 seconds.

Transverse Data

temperature will affect the mechanical properties of the steel. Similarly, the rate of cooling from the brazing temperature may have an effect on the final properties. These factors have been taken into account in developing braze-cycle heat treatments (BCHT). The required conditions for brazing have been combined with heating, cooling, refrigerating and aging steps, modified from the standard treatments where necessary, to develop adequate mechanical properties in the brazed assembly. Some of the cycles that have been used are fairly complex and the details are not included in this memorandum. However, it has been reported (Reference 30) that PH 14-8 Mo responds to a wide variety of braze-cycle heat treatments. The tabulated data on the yield strength of the alloy after being subjected to eleven different treatment cycles showed that the yield strength was well above the minimum acceptable for material specifications.

AM-350 and AM-355. These are controlled-transformation, precipitation-hardenable stainless steels that are dependent on thermal and mechanical treatments for their useful mechanical properties. Although there are some differences in the alloys which will be pointed out in the discussion, they are similar in many respects and it is appropriate to consider them simultaneously regarding their response to thermal and mechanical treatments.

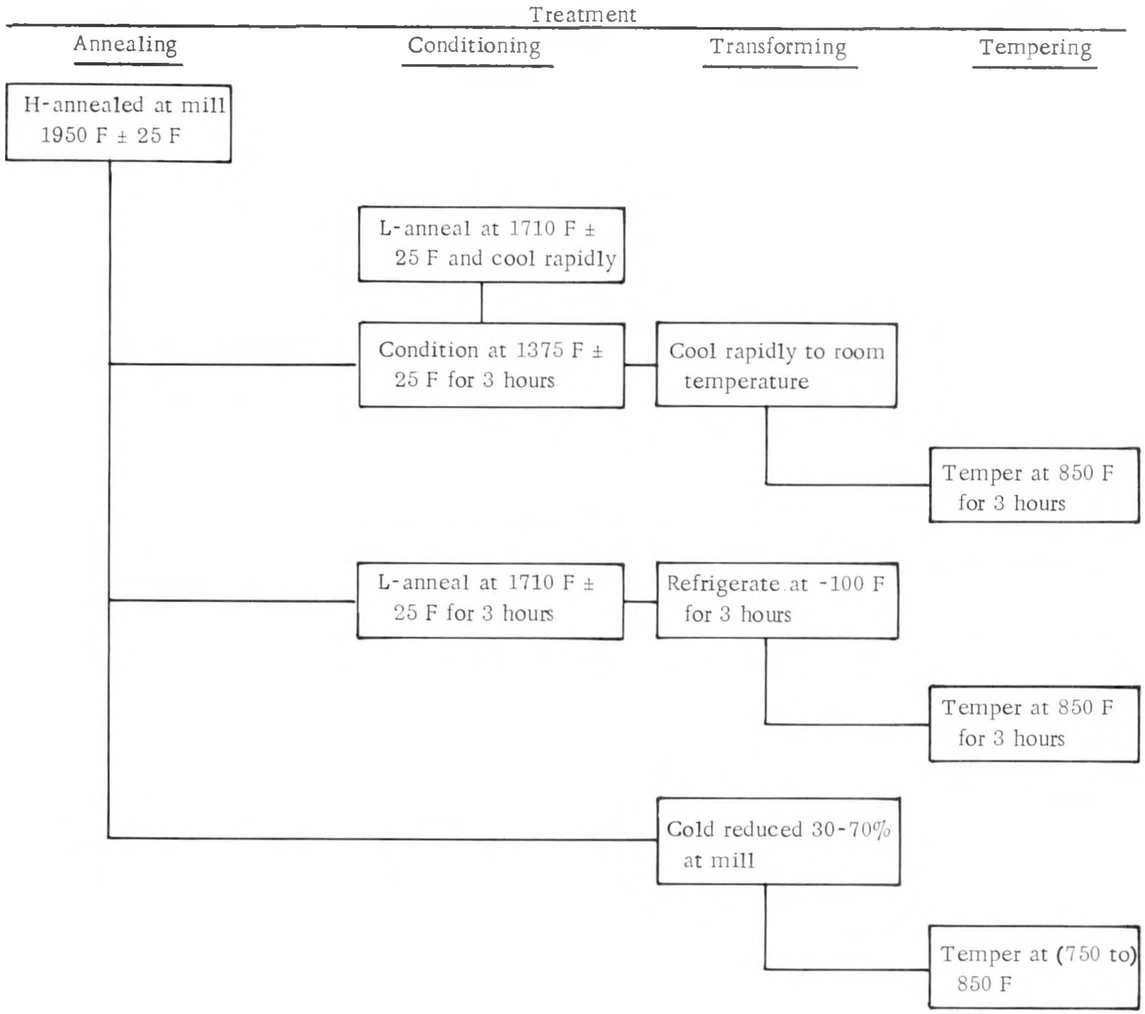
In the annealed condition, the alloys are austenitic and readily formable. Adequate strength and hardness are obtained by various sequences of conditioning, transformation, cold-rolling and tempering treatments. Considerable strengthening occurs by virtue of transformation of austenite to martensite, and additional increases in strength are possible by tempering the martensite. According to Lula (Ref. 31) the term "tempering" is not entirely descriptive because the additional increase in yield strength that is obtained by tempering is indicative of metallurgical changes such as precipitation-hardening. Recent research indicates that the mechanism of hardening during the tempering operation is by precipitation of an intermetallic compound tentatively identified as a chromium nitride.

As shown in Table I, AM 350 and AM 355 differ only in their carbon and chromium content. AM 355 has a slightly higher carbon and lower chromium content than AM 350. These relatively minor differences in composition are important, however, because they determine the structure and properties of the alloys that are developed by heat treatments, and also account for some differences in the treatment of the two alloys to obtain optimum properties.

The standard treatments for AM 350 are (1) subzero cooling and tempering (SCT), (2) double aging (DA), and (3) cold rolling and tempering. The various steps in these treatments are shown in Figure 50. The chart also gives typical tensile properties that are obtained following certain intermediate steps in the treatment sequences, as well as the properties in the standard final conditions. A similar chart for AM 355 sheet is shown in Figure 51. AM 355 is widely used in bar form which is treated somewhat differently from sheet material. The treatment sequences and typical properties obtained for AM 355 bar are shown in Figure 52. Castings are treated according to the schedules shown in Figure 53.

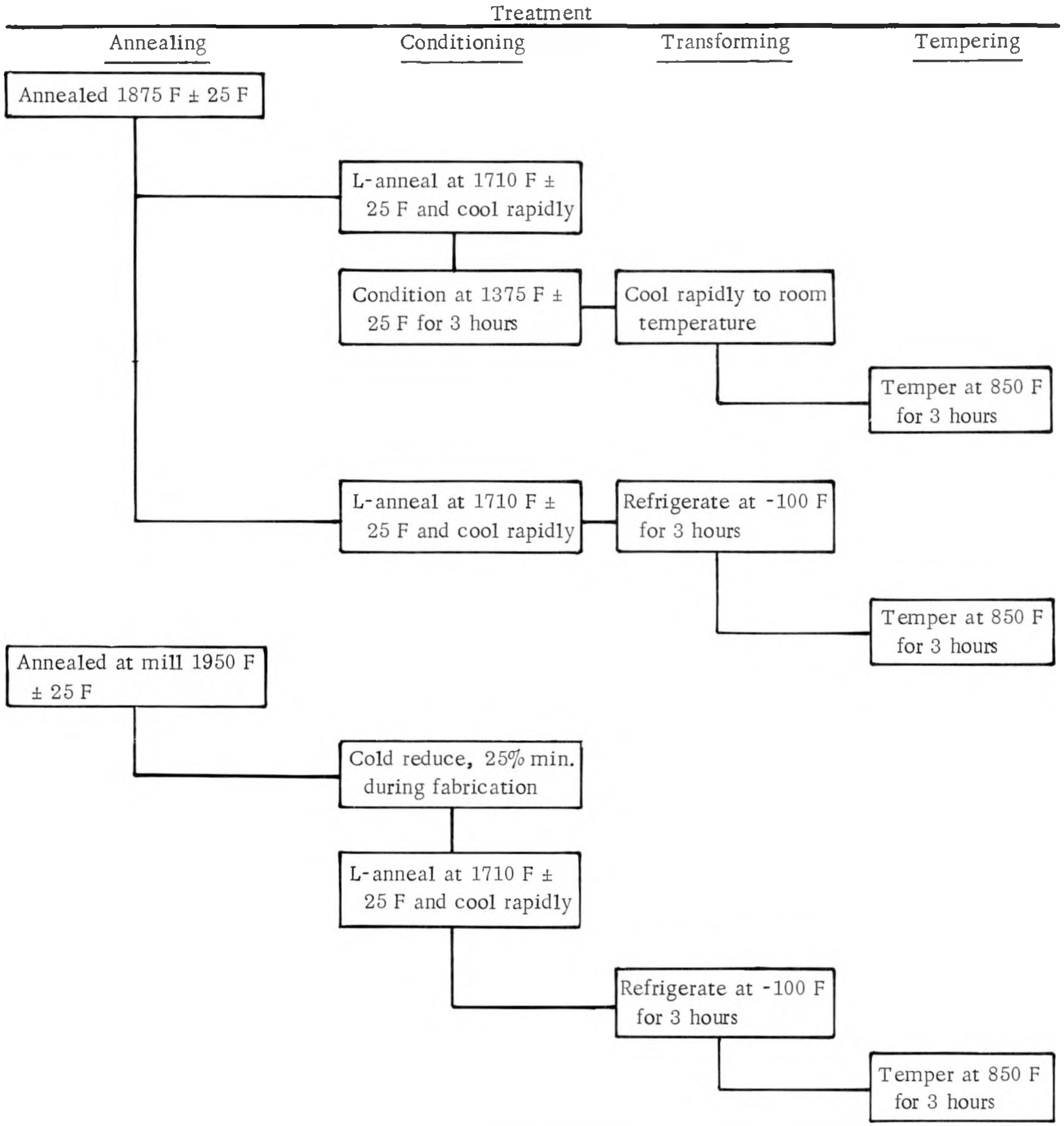
Additional comparisons of the properties of AM 350 and AM 355 in various standard forms and conditions are given in Tables XXXIX and XL. These show that the form of the alloy as well as the heat treatments determine the mechanical properties obtained.

Annealing. There is a marked difference in annealing behavior between AM 350 and AM 355 which results from the small difference in their carbon and chromium content. The ferrite-austenite balance is displaced so that the AM-355 structure shows little or no delta ferrite, while AM-350 may have as much as 10 to 20 percent of this constituent present. The presence of delta ferrite properly dispersed throughout the structure serves two purposes: (1) it restricts grain growth during high temperature (1900-1975 F) annealing, and (2) the delta ferrite-austenite interfaces serve as the preferred sites for carbide precipitation during the trigger-annealing (conditioning) treatment. When little or no delta ferrite is present, high annealing temperature will cause grain growth, and the conditioning treatment will result in precipitation of the carbides mainly at the grain boundaries. Transformation to martensite during subsequent treatment will be confined to the depleted areas adjacent to the grain boundaries.



Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation, per cent	Designation
149	63	39	H-annealed
166	56	20	L-annealed
160	105	10	
187	156	13.5	L + DA
186	149	12.5	H + DA
200	110	12	
200	170	13	SCT (850)
228	200	15	CR (30)
275	250	13	CR (50)
339	322	11	CR (70)
227	214	18	CRT (30)
272	259	12	CRT (50)
335	322	11	CRT (70)

FIGURE 50. TREATMENT AND TYPICAL PROPERTIES OF AM 350 (Ref. 2)



Typical Properties			
Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation, per cent	Designation
187	56	29	Annealed
195	60	15	L-annealed
195	155	10	DA
	130		
215	180	11	SCT
165	52	40	Annealed
225	195	13	SCT

FIGURE 51. TREATMENT AND TYPICAL PROPERTIES OF AM 355 SHEET (Ref. 2)

Treatment: Annealing	Treatment: Conditioning	Treatment: Transforming	Treatment: Tempering	%CR	Typical Properties: Tensile Strength, 1000 psi	Typical Properties: Yield Strength, 1000 psi	Typical Properties: Elongation, per cent	Designation
Mill annealed at 1900 to 1950 F								
		Cold rolled 25-50% at mill	Temper at 750 F- 850 F	20	210	175	23[a]	CRT
				30	240	220	18[a]	
				40	265	260	10[a]	
				20	215	165	20[b]	
				30	240	200	16[b]	
				40	265	235	8[b]	
		Cold rolled more than 50% at mill	Temper at 750 F - 850 F		350	335	1	XH[b]

(a) Longitudinal
(b) Transverse

FIGURE 51. (CONTINUED)

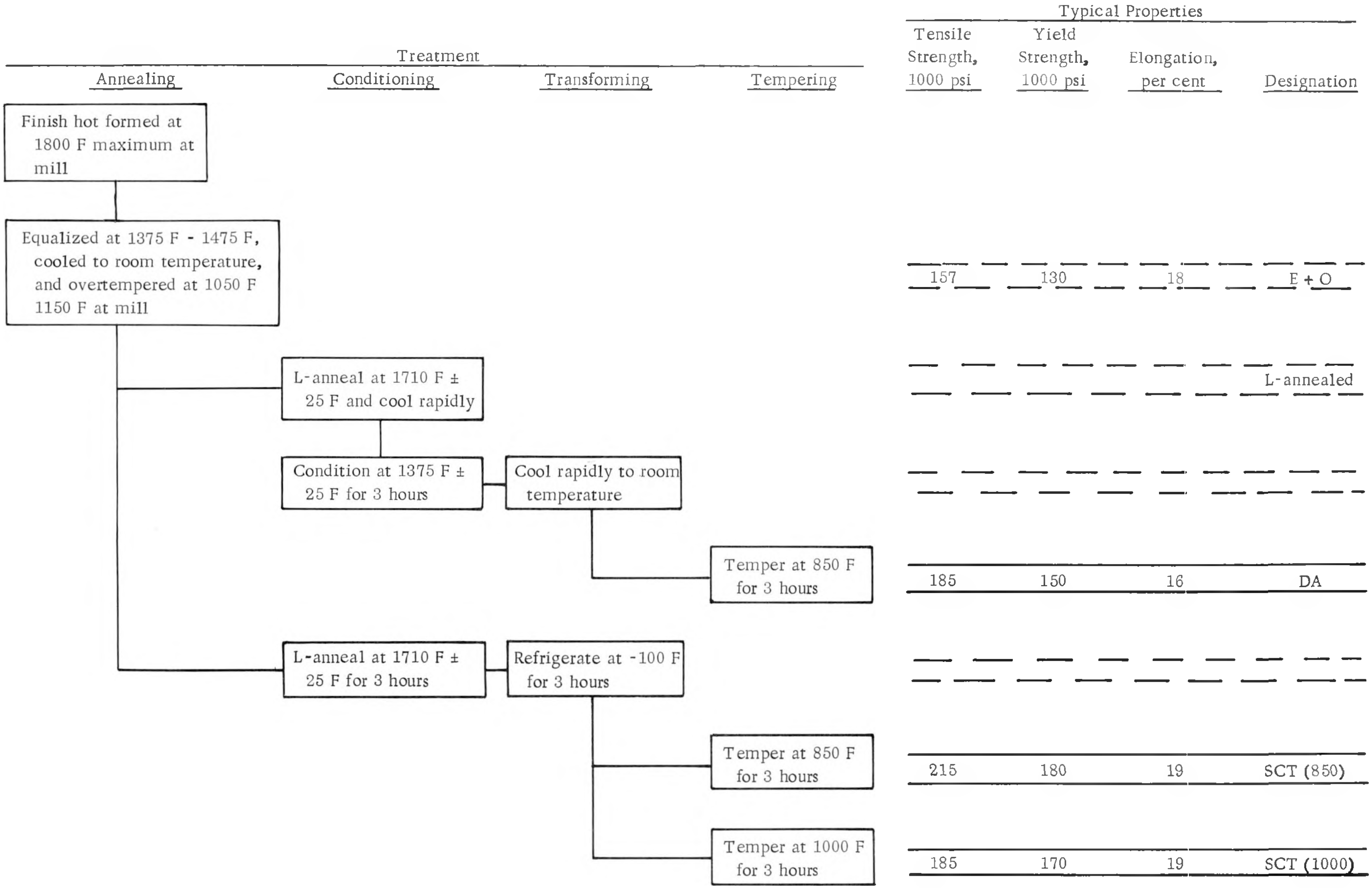


FIGURE 52. TREATMENT AND TYPICAL PROPERTIES OF AM 355 BAR (Ref. 2)

Treatment: Annealing	Treatment: Conditioning	Treatment: Transforming	Treatment: Tempering	Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation, per cent	Designation
Anneal at 2000 F ± 25 F				—	—	—	Annealed
Equalize at 1375 F - 1475 F, cool to room temperature, over temper at 1050 F - 1150 F				—	—	—	E + O
	L-anneal at 1750 F ± 25 F and cool rapidly			—	—	—	L-annealed
	Condition at 1375 F ± 25 F for 3 hours	Cool rapidly to room temperature		—	—	—	—
			Temper at 850 F for 3 hours	180	150	12	DA
	L-anneal at 1750 F ± 25 F and cool rapidly	Refrigerate at -100 F for 3 hours		—	—	—	—
			Temper at 850 F for 3 hours	215	169	12	SCT (850)
			Temper at 1000 F for 3 hours	174	140	16	SCT (1000)

FIGURE 53. TREATMENT AND TYPICAL PROPERTIES OF AM 355 CASTINGS (Ref. 2)

TABLE XXXIX. TYPICAL ROOM TEMPERATURE PROPERTIES OF AM 350 SHEET IN SEVERAL STANDARD CONDITIONS (REF. 32)

Condition	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elong. in 2 in., %	Hardness, Rockwell C
H	60	145	40.0	20.0
SCT (850 F)	173	206	13.5	45.0
SCT (1000 F)	148	169	15.0	38.0
L + DA	156	187	13.5	42.0
H + DA	149	186	12.5	41.5

TABLE XL. TYPICAL ROOM TEMPERATURE TENSILE PROPERTIES OF AM 355 IN SEVERAL STANDARD CONDITIONS (REF. 32)

Form	Condition	Yield Strength 0.02% Offset, ksi	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elong. in 2 in., %	Red. in Area, %
Bar	SCT (850 F)	142	182	216	19.0	38.5
	(1000 F)	147	171	186	19.0	57.0
Plate	H	-	57	160	26.0	-
	SCT (850 F)	130	174	210	14.0	40.0
	(1000 F)	132	158	175	17.0	48.0
Castings	SCT (850 F)	115	165	215	15.0	37.5
	(1000 F)	110	135	170	15.0	45.2
	DA	100	145	185	13.0	27.0

The influence of the temperature of annealing (of cold rolled material) on the strength of conditioned and hardened AM 350 and AM 355 is illustrated by Figure 54. This shows that AM 355 is stronger than AM 350, provided that the annealing temperature is kept below about 1875 F. Above that, the properties of AM 355 fall off rapidly while those of AM 350 remain essentially the same as at lower annealing temperature. Photomicrographs in the original publication show the carbides concentrated at the boundaries of the large grains developed during the 1950 F annealing of AM 355, and the more uniform structure resulting from the 1850 F annealing temperature. In the case of AM 350, there was no difference in the structure at the two annealing temperatures. Therefore, AM 350 sheet may be annealed at 1950 F but a temperature of about 1875 F should be used for AM 355 sheet in order to retain its strength advantage over AM 350.

The beneficial effect of delta ferrite indicated by the above results are only obtained if the delta ferrite is well-dispersed through the cross-section of the material. Proper dispersion is only obtained by hot or cold working of the alloy during mill production of sheet and small diameter bars, and for this reason AM 350 is not usually recommended for applications requiring heavy sections (Ref. 32). Conversely, the lack of delta ferrite in AM 355 is a handicap in sheet applications where maximum formability obtained by high temperature annealing is desired. For this reason, AM 355 finds its greatest application in bar and plate form.

Hardening Procedures. The conditioning, transformation and tempering procedures for AM 350 sheet are fairly straightforward and have been summarized in Figure 50.

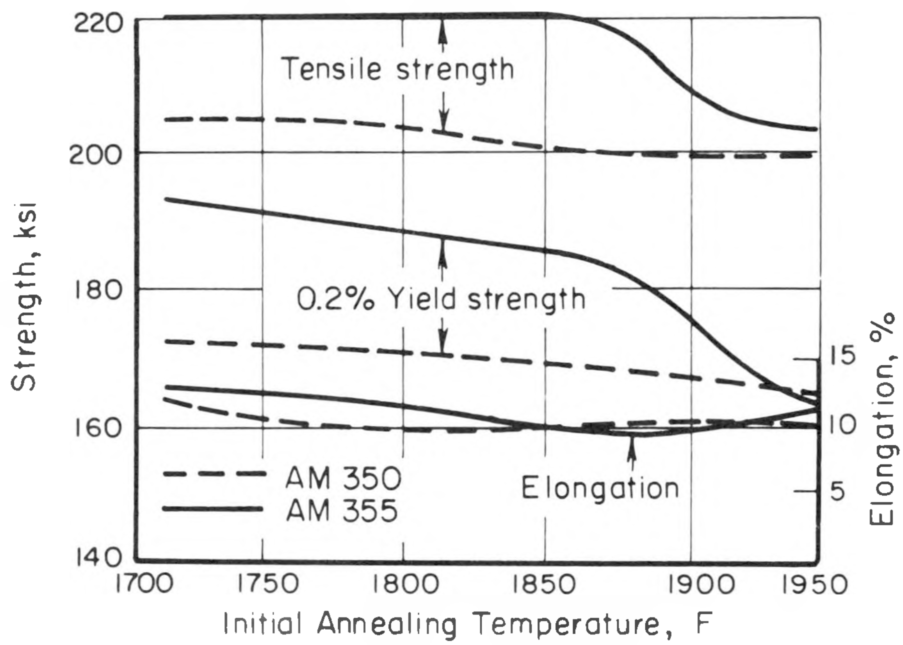


FIGURE 54. EFFECT OF ANNEALING ON THE S.C.T. PROPERTIES OF AM 350 AND AM 355 (Ref. 31)

The DA condition noted in the chart shows that the double aging treatment may be preceded by either the H-anneal or the L-anneal. These are differentiated by the designation H + DA and L + DA, respectively, and the properties obtained are listed in both Figure 50 and Table XXXIX. It is apparent that the prior annealing temperature has only a minor effect on the strength in the DA condition.

It has been indicated that the 1710 F conditioning treatment may be omitted, and that it is sufficient to condition at 1375 F only (Ref. 2). However, the dual conditioning results in more uniform and complete precipitation of carbides, and the material (after tempering) is about 5,000 psi stronger than material given only the 1375 F conditioning treatment.

The subzero-cooled and tempered condition, SCT, is sometimes designated as SCT (850) or SCT (1000), the numbers indicating the final tempering temperature. Tempering at 850 F will result in maximum hardening; but somewhat greater ductility, with a corresponding decrease in strength, may be obtained by tempering at the higher temperature. The properties obtained are listed in Table XXXIX. Additional data showing similar results for AM-350 bar stock are listed in Table XLI.

In the case of AM 355, several procedures have been suggested to overcome the tendency for intergranular carbide precipitation during conditioning heat treatments. One investigation was specifically concerned with improvements for sheet material, and the results were summarized by Aggen and Kaltenhauser (Ref. 33). The procedures that were studied in the sheet investigation involved additional steps called "preliminary heat treatment" inserted between the 1950 F anneal and the 1710 F conditioning treatment in the standard SCT sequence. The effect of these pretreatments on the mechanical properties of AM-355 (SCT) sheet are as follows:

TABLE XLI. EFFECT OF TEMPERING ON TENSILE PROPERTIES OF AM-350 BAR STOCK (REF. 32)

Tempering Temperature, F	Yield Strength 0.2% Offset, ksi	Ultimate Tensile Strength, ksi	Elong. in 2 in., %	Red. in Area, %	Hardness, Rockwell C
850	162	198	15	49	47
1000	150	164	22	53	40
1100	108	151	20	50	35

Condition: SCT (850, 1000 and 1100 F)

Preliminary Heat Treatment*	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in., %
None	156.8	204.6	17.5
1375 F, 3 hr	167.3	217.8	15.0
1710 F and sub-zero cool	183.3	222.2	13.5
Sub-zero cool	184.2	222.9	13.5

*Applied after 1950 F anneal and before 1710 F conditioning in the SCT (850) sequence.

The improvement provided by the last two pretreatments is quite marked, the properties being definitely better than those of normal AM 350 (SCT), and equivalent to those obtained on AM 355 when using an 1850 F annealing temperature in the SCT cycle. The improvement was correlated with the more uniform carbide precipitation within the grains. The martensite which was formed as a result of the pretreatments served as the sites for carbide precipitation during the 1710 F conditioning treatment.

A similar treatment has also been reported to improve carbide distribution in bar stock. It involves annealing at 1900 F followed by cooling rapidly enough to avoid intergranular carbide precipitation, refrigerating at -100 F, and then applying the conditioning, refrigerating and tempering treatments used for the standard SCT (850 or 1000). A marked improvement in impact strength was obtained as shown by Table XLII and Figure 55.

TABLE XLII. COMPARATIVE ROOM-TEMPERATURE CHARPY V-NOTCH IMPACT PROPERTIES OF AM-355 (REF. 33)

Heat Treatments: A—1750 F, 1 hr, water quenched + 3 hr at −100 F + 1000 F, 3 hr, air cooled.
B—1900 F, 1 hr, water quenched + 3 hr at −100 F + above heat Treatment A.
C—Same as A except tempered 3 hr at 850 F instead of 1000 F.
D—Same as B except tempered 3 hr at 850 F instead of 1000 F.

Product	Test Direction[a]	Heat Treatment	Charpy V-Notch Impact Energy, ft-lb
1⅛ in. dia	L	A	23.0
	L	B	31.0
1 in. dia	L	A	34.2
	L	B	41.5
1 in. plate	T	A	15.2
	T	B	19.7
4¼ in. square	T	A	5.0
	T	B	12.0
5 in. dia	T	A	5.8
	T	B	7.8
1¼ in. dia	L	A	19.8
	L	B	50.0
1¼ in. dia	L	C	13.3
	L	D	17.8
2¼ in. dia	L	A	33.3
	L	B	41.0
	T	A	8.3
	T	B	12.3

[a] L = longitudinal; T = transverse.

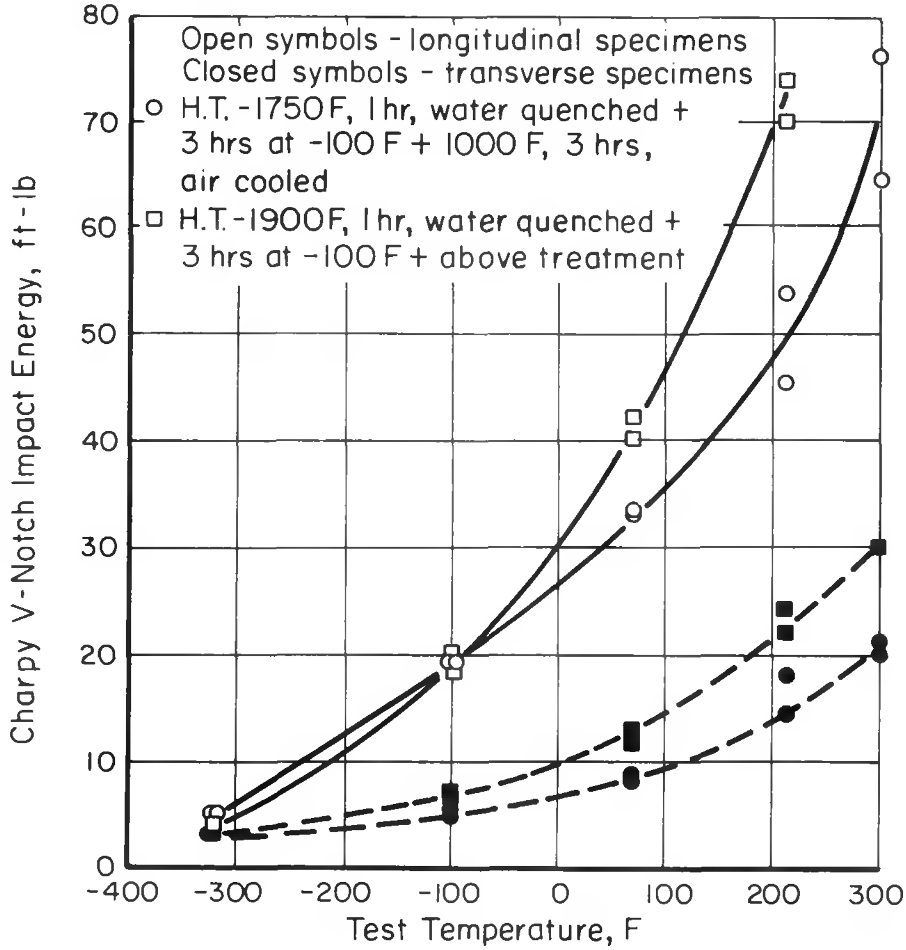


FIGURE 55. COMPARATIVE IMPACT ENERGY CURVES FOR AM-355 2-1/4 INCH-DIAMETER BAR (Ref. 33)

An improved structure can also be obtained by eliminating the 1950 F anneal and using final hot working operations involving substantial reduction (20 percent or more) at temperatures below 1875 F. This will result in a finer grain size and more grain boundary area to serve as sites for carbide precipitation during L-annealing. Martensitic areas resulting from cooling to room temperature after hot working also serve as sites for carbide precipitation.

Increased Charpy V-notch strength may be obtained by tempering at higher temperature in the standard SCT treatment. This is shown in Figure 56. The improved notch strength is accompanied by lower tensile and yield strengths.

Cold Working. In common with other semiaustenitic PH stainless steels, AM 350 and AM 355 may be hardened to very high strength levels by cold working. In the process, austenite is transformed to martensite. The factors that determine the strength produced have been reviewed and summarized by McCunn, et al (Ref. 35). The strength level with a given amount of working will increase with the amount and hardness of the martensite produced. The composition of the alloy, heat treatments prior to working, the rolling temperature, and amount of reduction, can all affect the extent of transformation, while the carbon content determines the hardness of the martensite that is formed.

AM-350 is somewhat more stable than AM-355 and also has a lower carbon content. Consequently, AM 350 develops somewhat lower mechanical properties than AM 355. The properties obtained by standard cold working and tempering treatments are given in Figure 50 for AM 350. Three degrees of cold reduction 30, 50, and 70 percent are compared. For AM 355 sheet, similar data are shown in Figure 51. Cold reductions of over 50 percent are called extra hardening, and designated as AM 355 (XH).

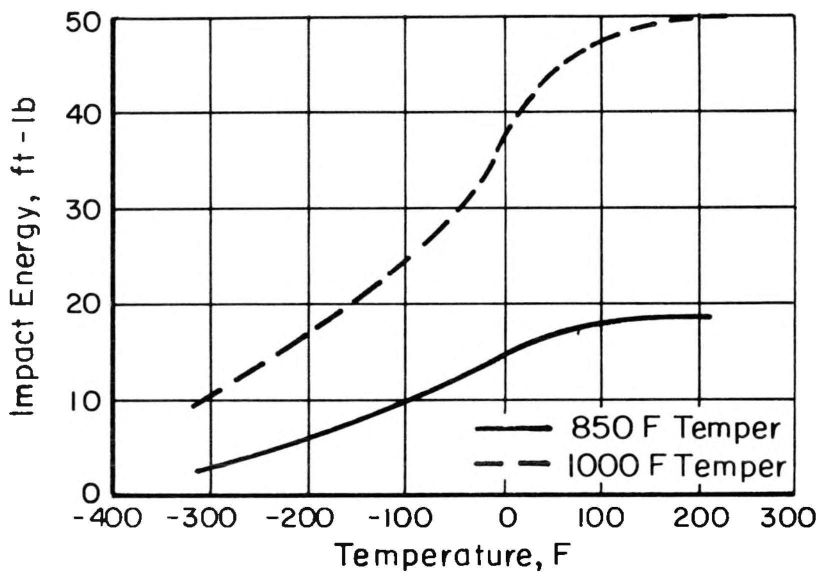


FIGURE 56. EFFECT OF TEMPERING TEMPERATURE ON CHARPY V-NOTCH STRENGTH OF AM-355, SCT (Ref. 34)

Longitudinal direction

For both alloys, the data in the figures show the substantial increase in tensile and yield strengths obtained with increasing deformation. A tempering treatment following cold rolling results in slightly increased yield strength for AM-350.

Considerable information on the properties obtained by various combinations of cold rolling and aging conditions was accumulated in the NASA evaluation of materials for the supersonic transport (Ref. 25). Table XLIII summarizes the data showing the effect of variations in these parameters on the room temperature tensile properties of AM-350 sheet. Table XLIV compares the smooth tensile strength with notched tensile strength obtained as a result of the changes in cold rolling and aging conditions. The results indicate that the optimum aging temperature depends on the amount of cold work given the sheet. Cold-worked responds to aging at a lower temperature than that transformed by conditioning and subzero cooling. The results indicate that some overaging occurred on specimens aged at 950 F.

Additional data comparing the properties obtained by different amounts of cold work on AM 355 to supplement the information in Figure 51 is given in the following tabulation (Ref. 35):

Condition	Test Direction	0.2% Yield Strength, ksi	Ultimate Tensile Strength, ksi	Elongation, %
CRT	longitudinal	220	240	17.0
	transverse	200	240	15.0
XH	longitudinal	340	342	1.0
	transverse	335	360	1.0

TABLE XLIII. EFFECT OF COLD-ROLLING AND AGING ON THE ROOM TEMPERATURE TENSILE PROPERTIES OF 0.025-INCH AM-350 SHEET (REF. 25)

Cold Rolling and Aging Conditions	Ultimate Tensile Strength, ksi		Yield Strength 0.2% Offset, ksi		Elongation, in 2 inches, %	
	L(a)	T	L	T	L	T
20% CR, 3 hr 950 F	198.8	205.8	168,0	167.7	19	17
20% CR, 3 hr 825 F	223.7	218.6	179.4	171.1	24.5	24.5
20% CR, 3 hr 850 F	219.9	-	186.2	-	22.7	-
30% CR, 3 hr 950 F	208.7	215.7	204.1	202.1	16.5	13.5
30% CR, 3 hr 825 F	245.0	243.8	239.0	221.9	20.0	15.0
30% CR, 3 hr 700 F	242.5	246.7	240.6	228.4	9.0	9.0
45% CR, 3 hr 825 F	280.0	280.0	274.2	274.1	(b)	3.5

(a) L = longitudinal, T = transverse.

(b) Broke outside gauge marks.

TABLE XLIV. EFFECT OF COLD ROLLING AND AGING ON THE ROOM TEMPERATURE NOTCH STRENGTH OF 0.025 INCH AM-350 SHEET (REF. 25)

Cold Rolling and Aging Conditions	Ultimate Tensile Strength, ksi		Notched Tensile Strength[a]		N/S Strength Ratio[b]	
	L[c]	T	L	T	L	T
20% CR, 3 hr 950 F	198.8	205.8	194.0	191.0	0.98	0.93
20% CR, 3 hr 825 F	223.7	218.6	216.2	203.7	0.97	0.93
20% CR, 3 hr 850 F	219.9	-	214.6	-	0.98	-
30% CR, 3 hr 950 F	208.7	215.7	228.3	218.4	1.09	1.01
30% CR, 3 hr 825 F	245.0	243.8	256.5	229.7	1.04	0.94
30% CR, 3 hr 700 F	242.5	246.7	258.7	231.3	1.07	0.94
45% CR, 3 hr 825 F	280.0	280.0	274.7	230.0	0.98	0.82

(a) Sharp edge-notches; notch radius <0.0007.

(b) Notched-to-smooth tensile strength ratio.

(c) L = longitudinal, T = transverse.

Two of the factors affecting the strength of cold rolled AM-355 mentioned earlier are the amount of cold reduction and heat treatment prior to working. The result of changes in both of these parameters is illustrated in Figure 57. All the curves show increasing yield and tensile strength, and decreasing elongation, with increasing amounts of cold work. However, the influence of prior annealing temperature on the stability of the austenite is indicated by the difference in the slopes of the yield strength curves. The curve for material annealed at 1710 F prior to cold rolling rises very steeply with relatively small amounts of cold work, indicating substantial transformation to martensite. Transformation is completed by only about 10 percent of cold reduction. On the other hand, the more stable austenite obtained on annealing at 1900 F requires considerably more cold working (50 percent or more) to be transformed to its maximum potential. The increase in strength beyond the point of nearly complete transformation for both annealing conditions is attributed to work hardening of the martensite (Ref. 35). The effect of the carbon content is shown by the higher strength of the transformed material following the 1900 F anneal. This occurs because more carbon is dissolved in the original austenite by higher temperature annealing, and this is retained in the martensite formed by cold rolling.

The increase in strength obtained in AM 350 by cold working (stretching), combined with the SCT(850) sequence is shown in Figure 58.

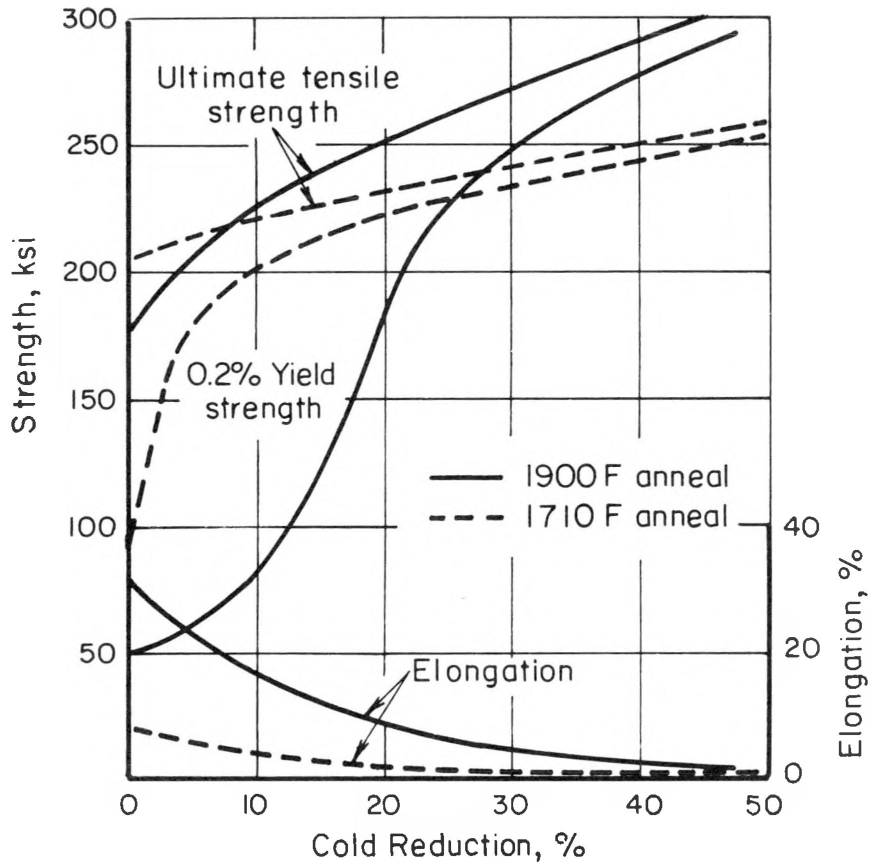


FIGURE 57. EFFECT OF COLD REDUCTION AND ANNEALING ON LONGITUDINAL TENSILE PROPERTIES OF AM 355 (Ref. 35)

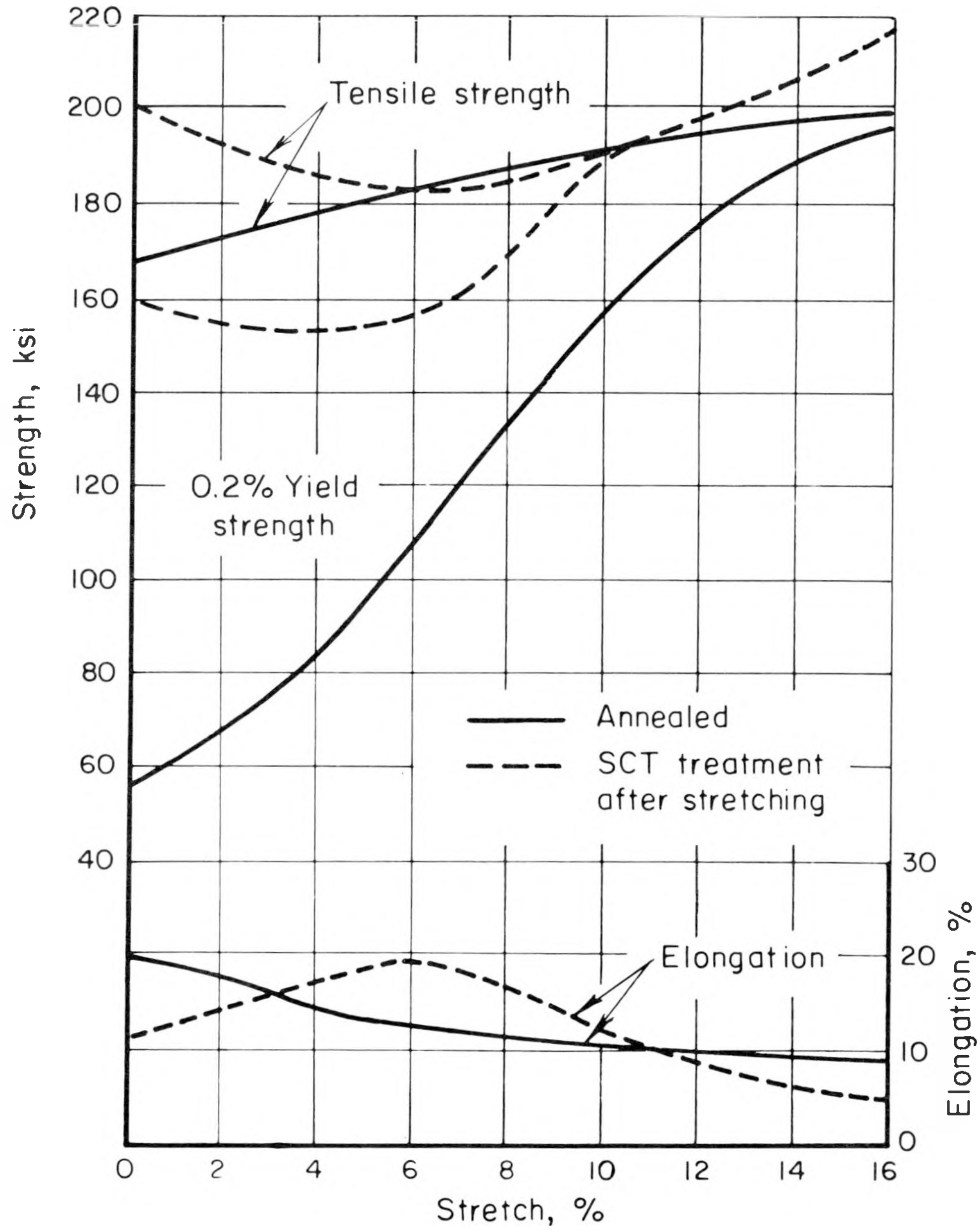


FIGURE 58. EFFECT OF STRETCHING ON THE TENSILE PROPERTIES OF AM 350 ANNEALED AT 1750 F (Ref. 36)

Warm Working. AM 350 can be formed at slightly elevated temperature (300 F) with little or no transformation of austenite (Ref. 35). At this temperature work hardening is low, permitting more severe forming operations. The effect of rolling temperature on the tensile properties of AM 355 is shown in Figure 59. This shows that the tensile properties decrease as the working temperature is increased to about 300 F. At higher temperatures no further changes in properties are obtained indicating that the Md temperature (the temperature above which transformation does not take place) has been reached.

Stress aging at elevated temperature was discussed in the section on 17-7PH (page 120). The treatments applied to AM-355, DA and the results obtained are shown in Figure 60.

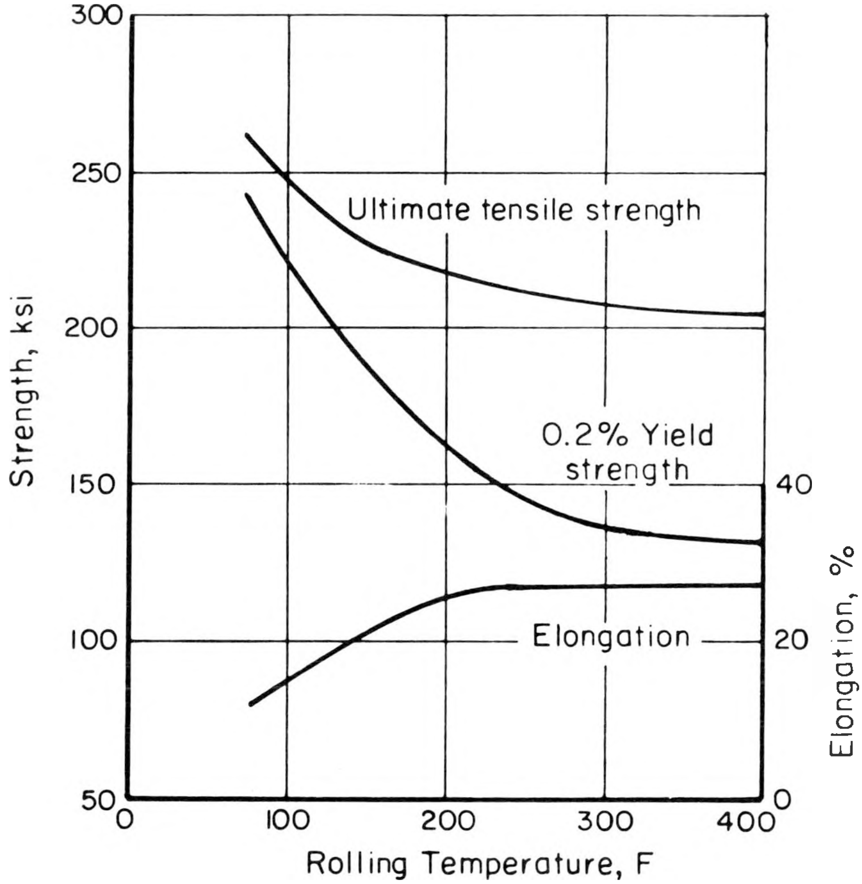


FIGURE 59. EFFECT OF ROLLING TEMPERATURE ON LONGITUDINAL TENSILE PROPERTIES OF AM 355 (Ref. 35)

Annealed at 1950 F, cold rolled 30 percent at indicated temperature, tempered 3 hours at 825 F

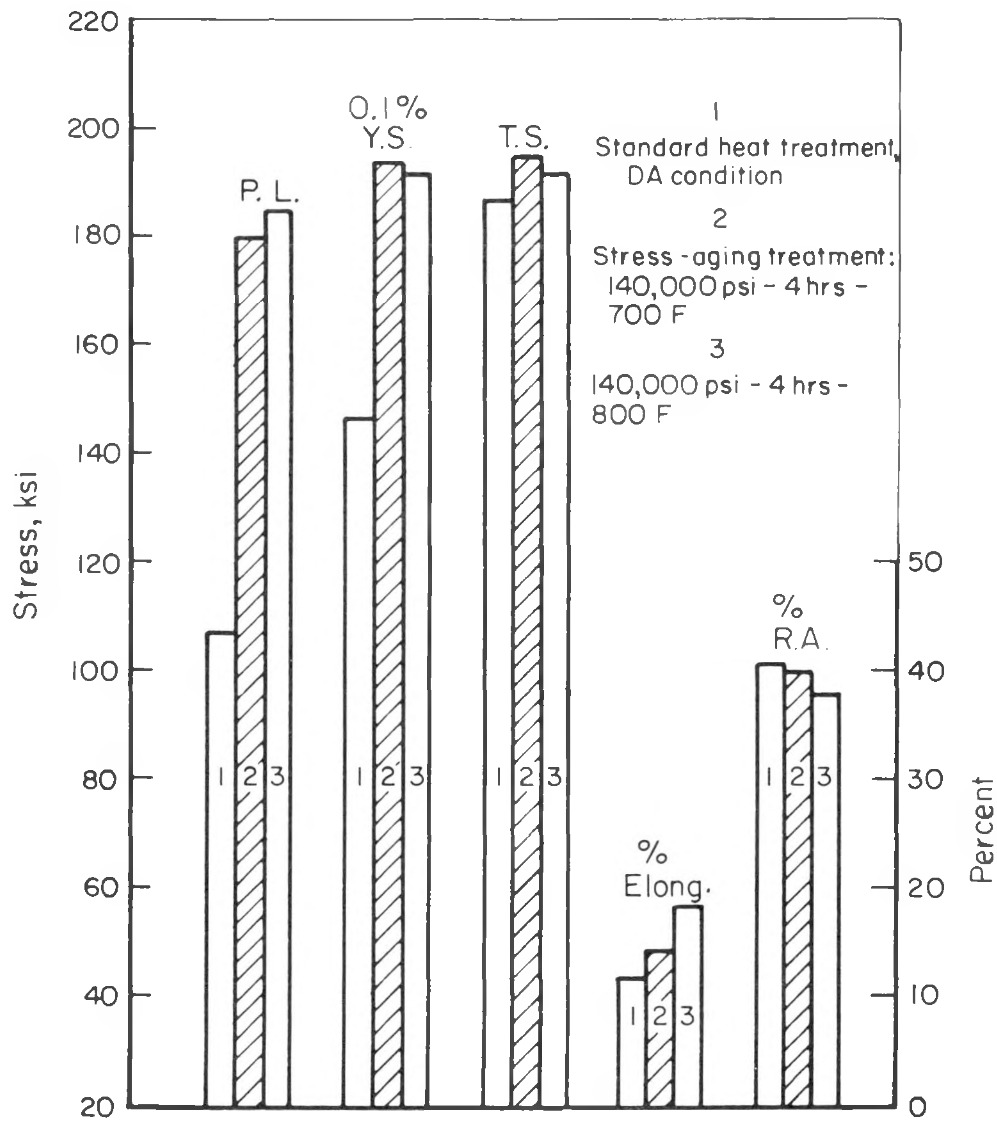


FIGURE 60. EFFECT OF STRESS-AGING TREATMENTS APPLIED TO AM 355 BAR STOCK IN THE DA CONDITION (Ref. 13)

AUSTENITIC PH STAINLESS STEELS

A-286. This is a relatively highly alloyed steel containing sufficient nickel to favor the austenitic structure and to reduce the Ms temperature so that the austenite is retained on cooling to room temperature as well as to cryogenic temperatures after annealing. The alloy contains titanium and aluminum and is hardened by the precipitation of an intermetallic compound Ni_3 (Al, Ti) in the austenite during an aging heat treatment.

The heat treatment for the alloy is relatively simple. It is solution annealed at either 1800 ±25 F or 1650 ±25 F for 90 minutes per inch of thickness, cooled rapidly, and aged at 1325 ±25 F for 16 hours and air-cooled. The properties obtained on bar stock by the standard treatment are as follows (Ref. 37):

	Solution Annealed at 1800 F	Aged at 1325 F
Ultimate tensile strength, ksi	91	146
0.2% offset yield strength, ksi	37	100
Elongation in 2 inches, percent	47.5	25

Selection of the solution-annealing temperature depends on the properties desired. Better creep and rupture strengths may be obtained by annealing at 1800 F, while annealing at 1650 F is reported to give better short-time tensile properties. The effect of annealing conditions on the properties of A-286 forgings is shown in Table L. In these tests the annealing conditions had little effect on the yield and tensile strengths, but improved room temperature ductility and stress rupture properties at 1200 F were obtained by annealing at 1800 F prior to aging at 1350 F.

TABLE XLV. EFFECT OF ANNEALING PROCEDURE ON THE MECHANICAL PROPERTIES OF A-286 FORGINGS (REF. 38)

Tensile data at room temperature

Heat Treatment	Ultimate Tensile Strength, ksi	Yield Strength, 0.2% Offset, ksi	Elongation in 4 D, percent	Reduction of Area, percent	Stress Rupture Life at 1200 F and 65 ksi, hr
1650 F, 2 hr, OQ, 1350 F, 16 hr[(a)]	152.0	109.0	18.0	22.0	48
1800 F, 1 hr, OQ, 1350 F, 16 hr[(b)]	150.0	107.0	24.0	40.0	100

(a) Average of 10 tests from different forgings.

(b) Average of 8 different forgings.

In practical heat treatment operations, acceptable properties are not always obtained when using a standard annealing and aging sequence. It is often possible to correct such situations by re-aging the alloy at a different temperature, or by some other adjustment in the heat treatment conditions. This is shown in the following example for A-286 (Ref. 39).

Treatment	Yield Strength 0.2% Offset, ksi
Standard 1650 F, 2 hr., OQ 1300 F, 16 hr, AC	Heat 1 - 89.1 Heat 2 - 91.6
As above + second aging at 1200 F, 16 hr, AC	Heat 1 - 91.6 Heat 2 - 101.2

The effect of cooling rate on short-time tensile properties is shown in Table XLVI. Alloy 286 is relatively insensitive to the rate of cooling from the annealing temperature, and thus section thickness is not of major concern in relation to cooling practice.

A comparison of the properties of annealed and age-hardened A-286 sheet at room and cryogenic temperatures to -423 F is shown in Figure 61. The increased strength at room temperature provided by aging is maintained at subzero temperatures. Elongation is reduced by aging but remains at an acceptable level at all test temperatures. A-286 has useful properties for applications at very low temperatures.

The alloy in sheet form can be strengthened considerably merely by cold-working, but elongation is drastically reduced. This is shown in Table XLVII. The data indicate that A-286 work hardens rapidly. The elongation is reduced from about 40 percent in the annealed condition to 5 percent after 30 percent cold reduction. The tensile and yield strengths continue to increase with additional cold working to 80 percent reduction, but the elongation remains at a low level.

TABLE XLVI. EFFECT OF COOLING RATE FROM 1800 to 1000 F on SHORT-TIME PROPERTIES OF A-286[a] (REF 39)

Cooling rate, °F per min	0.2% yield strength, psi	Tensile strength, psi	Elongation, %	Reduction of area, %
Tested at 70 F				
Oil quench..	99,000	149,000	24.0	38.5
26.6	100,000	150,000	22.4	42.0
15.2	98,700	147,800	24.0	38.5
8.7	102,000	143,000	22.4	36.0
2.2	96,000	149,000	23.2	33.0
Tested at 1200 F				
Oil quench..	93,500	112,800	8.1	10.3
26.6	94,000	112,000	5.6	7.4
15.2	92,000	112,200	8.8	16.0
8.7	86,300	107,700	7.2	17.3
2.2	90,000	108,000	11.2	18.4

(a) After cooling from the solution temperature, all samples were aged at 1325 F for 16 hr and air cooled.

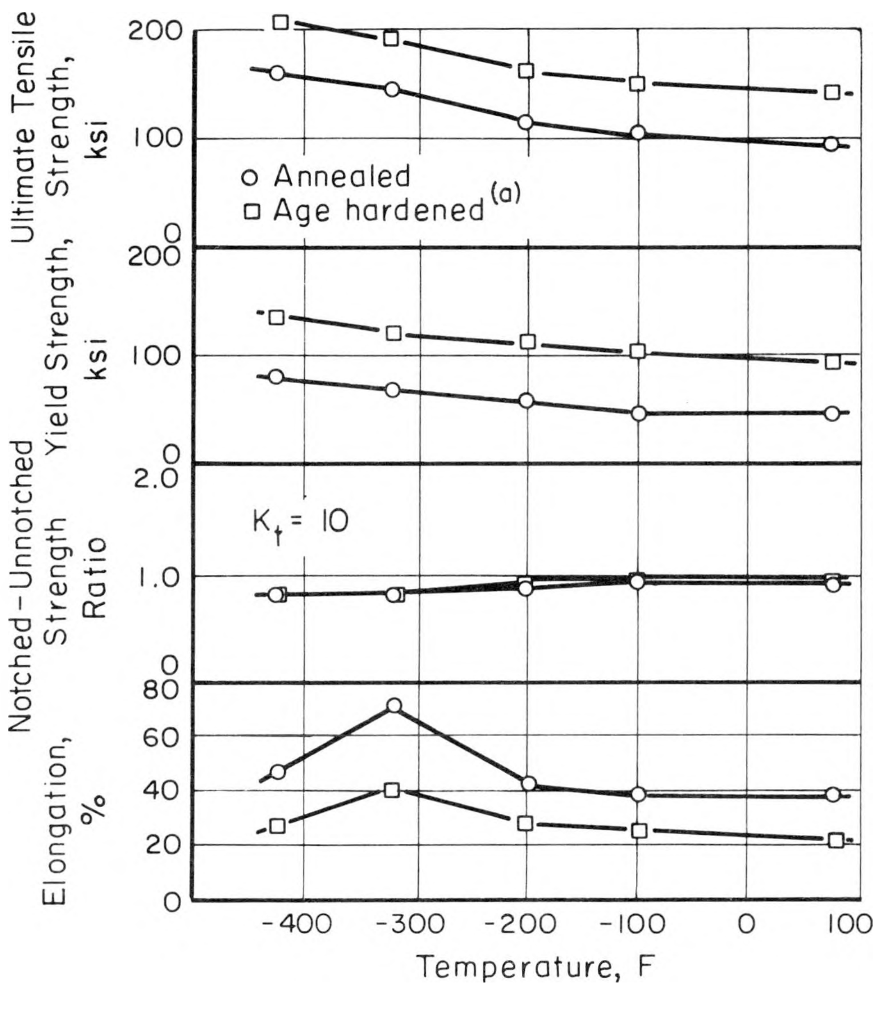


FIGURE 61. CRYOGENIC TEMPERATURE PROPERTIES OF A-286 SHEET. EFFECT OF AGE HARDENING. (Ref. 40)

0.060 and 0.095-inch sheet

(a) Age hardening - heat to 1350 F, hold 16 hours, air cool

TABLE XLVII. EFFECT OF COLD WORKING ON ROOM-TEMPERATURE TENSILE PROPERTIES OF A-286 SHEET (REFS 25, 37)

Tested after cold working only; not aged

Amount of Cold Reduction, %	Ultimate Tensile Strength, ksi	Yield Strength 0.2% Offset, ksi	Elongation in 2 inches, %
0	89	41.7	40.0
30	L[(a)] 132	120	5.0
30	T[(a)] 133	121	7.0
50	L 152	143	3.5
50	T 162	147	3.5
65	L 167	149	2.5
65	T 180	157	2.5
80	L 178	153	2.7
80	T 204	177	3.5

(a) L = longitudinal

T = transverse

Still further increase in strength is possible with a combination of cold working and age hardening. Information on the effect of various amounts of cold reduction followed by several different aging heat treatments was accumulated in the NASA evaluation of materials for the supersonic transport. The data are assembled in Table XLVIII.

It appears that severely cold-worked material precipitation hardens at a lower temperature than less severely worked products. There is an optimum aging temperature for maximum strengthening that depends on the extent of cold reduction. The greater the cold reduction, the lower the aging temperature necessary to develop the highest tensile strength. This is attributed to the greater energy available for precipitation during aging, in severely worked areas. At higher aging temperatures, tensile and yield strengths are decreased, and the elongation is increased indicating an overaged condition. This effect is also shown in Figure 62. illustrating the relationship between cold working and aging conditions on the hardness of A-286 sheet. Maximum hardness is developed by aging at temperatures that depend on the degree of prior cold work.

The effect of aging and cold working on the notched strength of A-286 sheet is shown in Table XLIX. Notched-unnotched tensile strength ratios are considerably lower on material that is cold worked more than about 50 percent.

TABLE XLVIII. EFFECT OF COLD WORK AND AGING ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF 0.025-INCH A-286 SHEET (REF. 25)

		CONDITION OF MATERIAL PRIOR TO AGING											
		COLD WORKED 30%			COLD WORKED 50%			COLD WORKED 65%			COLD WORKED 80%		
Heat Treatment	Direction[a]	UTS[b] (1000 psi)	YS (1000 psi)	Elong (%)	UTS (1000 psi)	YS (1000 psi)	Elong. (%)	UTS (1000 psi)	YS (1000 psi)	Elong. (%)	UTS (1000 psi)	YS (1000 psi)	Elong. (%)
None	L	132	120	5	152	143	3.5	167	149	2.5	178	153	2.7
	T	133	121	7	162	147	3.3	180	157	2.5	204	177	3.5
16 hrs at 1100°F	L	---	---	-	---	---	---	221	212	3	239	230	2.8
	T	---	---	-	---	---	---	---	---	---	263	250	3.8
16 hrs at 1200°F	L	183	169	9	205	192	6	217	204	3.8	224	202	3
	T	---	---	-	---	---	--	232	220	6	243	226	4.5
16 hrs at 1300°F	L	185	163	11	182	153	11	173	139	10	170	136	8
	L	---	---	-	187	153	10	177	140	10	174	135	10
	T	181	154	10	---	---	---	184	157	10	---	---	---

(a) L = Longitudinal

T = Transverse

(b) UTS - ultimate tensile strength; YS - 0.2 percent offset yield strength
Elong. - elongation in 2 inches.

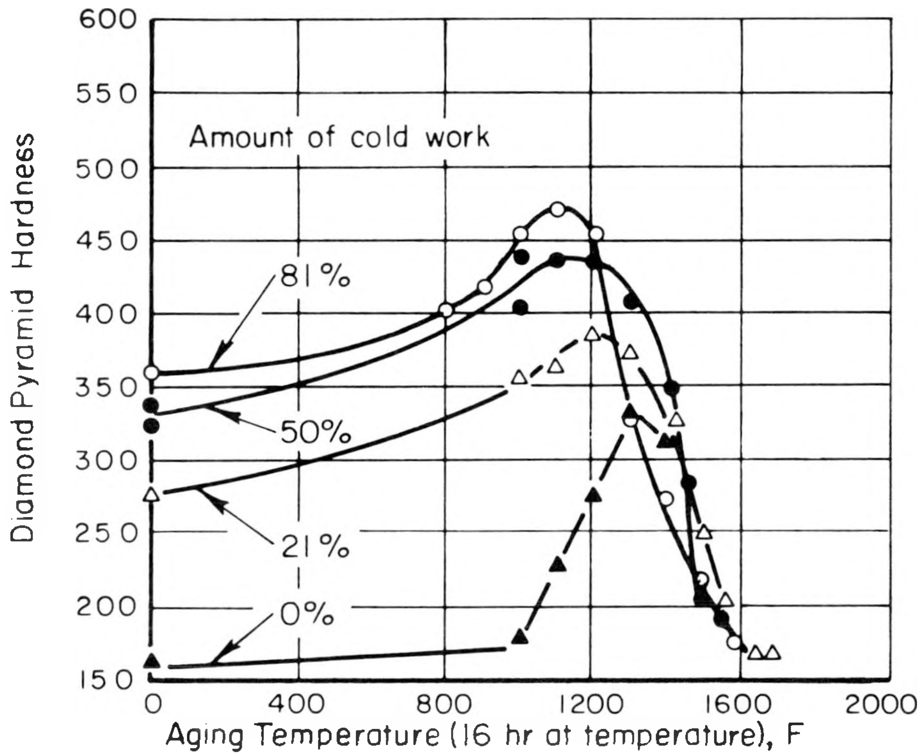


FIGURE 62. EFFECT OF COLD WORK AND AGING ON THE HARDNESS OF A-286 SHEET (Ref. 39)

Diamond pyramid hardness (10 kg-load) after cold working and aging 16 hours. Material solution treated at 1800 F for 1 hour and oil quenched before cold working and aging.

TABLE XLIX. NOTCHED/UNNOTCHED TENSILE STRENGTH RATIO OF A-286 SHEET AS INFLUENCED BY COLD WORK AND AGING (REF. 25)

0.025-inch sheet

Condition	Direction[a]	Ultimate Tensile Strength, ksi	Notched Tensile Strength, ksi	Notched/ Unnotched Ultimate Strength Ratio
CR 30%, 16 hr, 1300 F	L	185	180	0.97
	T	181	180	0.99
CR 65%, 16 hr, 1200 F	L	217	143	0.66
	T	232	146	0.63
CR 80%, 16 hr, 1100 F	L	239	143	0.60
	T	263	140	0.53
CR 80%, 16 hr, 1200 F	L	224	116	0.52
	T	243	109	0.45

(a) L = longitudinal

T = transverse

(b) ASTM edge-notch sample; notch root radius <0.0007.

(c) Notched-to-smooth tensile strength ratio.

The effect of cold working <u>after aging</u> on the properties of A-286 bolts was determined in experiments conducted by Montano (Ref. 41). Various tests were made and the results concerned with the effects of metal heat- or working-treatments are summarized in Table L. The results indicate that a better combination of cryogenic temperature properties are obtained by cold-working only 50 percent after aging. Cold working to a 65 percent reduction resulted in only slightly higher strength, but much poorer notch strength at temperatures below -200 F.

The hardening effect of shock waves on solution-treated A-286 is seen in Figure 63. The hardness was a function of the shock pressure applied to the solution-treated material. Aging experiments on specimens that had been subjected to shock waves showed that the aging response was a function of the intensity of the shock treatment. The change in hardness with time of aging for shocked and unshocked material is shown in Figure 64.

TABLE L. CRYOGENIC TEMPERATURE PROPERTIES OF A-286 BOLTS AS INFLUENCED BY DEGREE OF COLD-WORKING AFTER AGING (REF. 41)

Test Temperature, F	Condition	Ultimate[a] Tensile Strength, ksi	Yield Strength[a] 0.2% Offset ksi	Elongation[a] in 2 inches, percent	Notched[b] Tensile Strength, ksi	Notched/ Unnotched Ultimate Strength Ratio[c]
75	Aged, 50% CW	206.6	191.1	9.3	283.5	1.372
-100	Ditto	219.8	201.2	11.7	303.8	1.382
-200	"	232.2	210.7	11.8	309.7	1.334
-320	"	262.8	223.7	13.5	331.7	1.266
-423	"	280.1	244.1	14.0	335.8	1.199
75	Aged, 65% CW	223.0	206.0	5.0	280.2	1.257
-100	Ditto	238.0	218.0	6.0	292.1	1.227
-200	"	250.6	230.3	7.5	304.1	1.213
-320	"	277.6	231.8	10.8	302.1	1.088
-423	"	290.2	247.5	7.6	282.8	0.974

(a) Determined on bolts after machining a smooth reduced section in the shank.

(b) Determined on bolts with a V-notch (K_t = 10) machined into original shank.

(c) Notched-to-reduced shank tensile strength ratio.

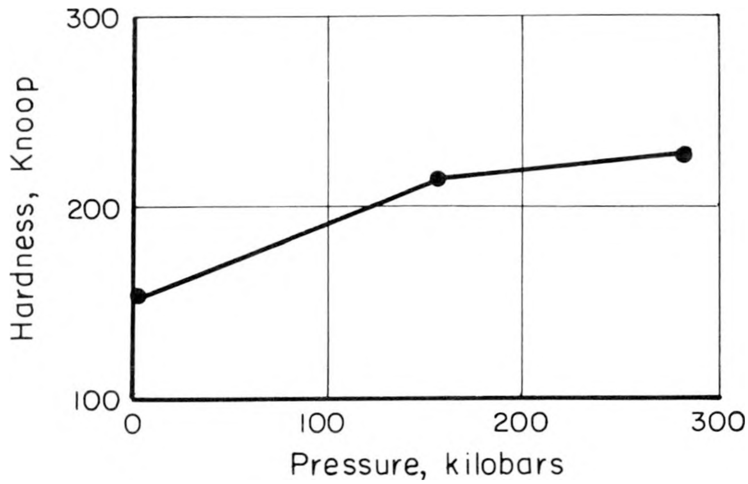


FIGURE 63. EFFECT OF SHOCK LOADING ON THE HARDNESS OF SOLUTION-TREATED A-286 (Ref. 42)

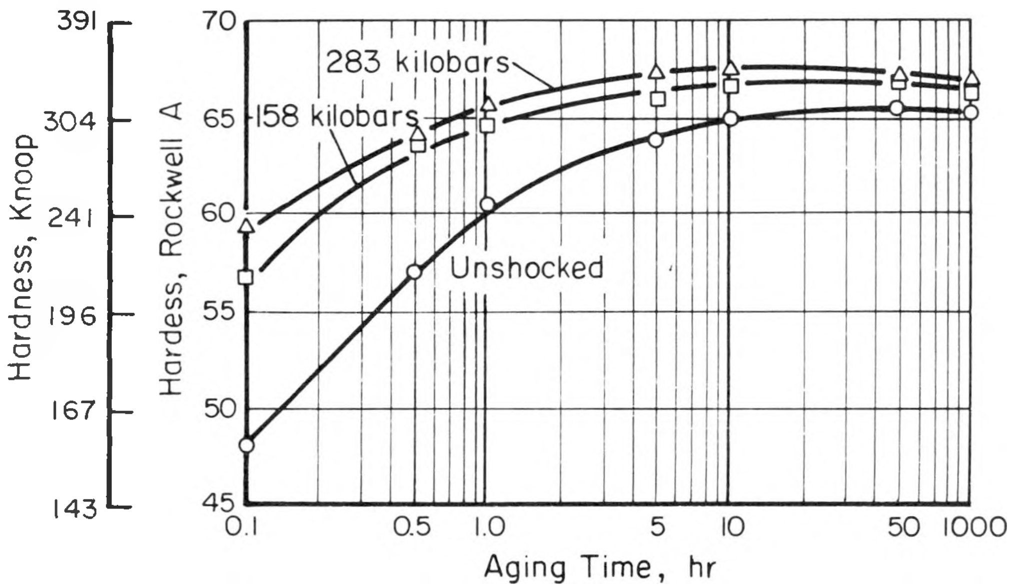


FIGURE 64. EFFECT OF SHOCK LOADING ON THE AGING RESPONSE OF A-286 (Ref. 42)

All Specimens Aged at 1315 F

CONCLUSIONS AND RECOMMENDATIONS

It is evident that since their emergence during World War II, the precipitation-hardenable stainless steels have become firmly established both as structural and as special purpose materials. The three types of PH stainless steel that have been developed; namely, martensitic, semiaustenitic, and austenitic, respond to a wide variety of thermal and mechanical treatments which exert great influence on their mechanical properties. As a result, these materials offer a tremendous range of mechanical properties in a wide spectrum of mill forms, semifinished and finished products.

Considerably more information is available regarding the effect of mechanical and thermal processing variables on the mechanical properties of such steels as 17-4PH, 17-7PH, PH15-7Mo, AM 350, AM 355, and A-286 than of 15-5PH, PH13-8Mo, AM 362, AM 363, and AFC 77. Such is to be expected because the former group is older and better established. In addition, a very good comprehension has been achieved of the physical metallurgy underlying the behavior of all three types of PH stainless steel.

While the PH stainless steels are exceedingly versatile and possess a well-earned reputation for satisfactory service, it is clear that they are specialty alloys. As such, close attention should be paid to the manner in which they are fabricated and heat treated; in general, procedures recommended by the producers should be followed if the great potentialities of these materials are to be realized. Deviations from standard practices should be made only after thorough investigation of the situation. Each step in the processing schedule is important; however, particular attention should be paid to such factors as heating atmosphere, temperature control during heating operations, cooling rates, hot working temperatures, the degree of final reduction in working and forming operations, and the quality and cleanliness of the surfaces of work pieces and parts.

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APPROVAL

THERMAL AND MECHANICAL TREATMENT FOR PRECIPITATION-HARDENING STAINLESS STEELS

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The information in this report has been reviewed for security classification. Review of any information concerning Department of Defense or Atomic Energy Commission programs has been made by the MSFC Security Classification Officer. This report, in its entirety, has been determined to be unclassified.

This document has also been reviewed and approved for technical accuracy.

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