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A REPORT

NATIONAL AERONAUTICS AND SPACE ADMINISTRATION

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THERMAL AND MECHANICAL TREATMENTS FOR NICKEL AND SOME NICKEL-BASE ALLOYS: EFFECTS ON MECHANICAL PROPERTIES

A REPORT

By
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Foreword

The Technology Utilization Office of the National Aeronautics and Space Administration has sponsored a series of state-of-the-art reports in the general field of materials fabrication. This is part of NASA's program to make generally available technological information resulting from its research and development efforts, which may have potential utility outside the aerospace community. The Columbus Laboratories, Battelle Memorial Institute, originally prepared these reports in 1965 and later revised them, updating the information to include the latest technology through 1968.

This report is one of a series pertaining to the fabricating of nickel, nickel-base, and cobalt-base alloys. Another series of volumes, in which the technology of processing precipitation-hardening stainless steels was surveyed, is available in NASA Special Publications, 5084 through 5090. Similar reports on low-alloy high-strength steels, stainless steel, and titanium alloys are planned.

This report deals with heat treating and working nickel and nickel-base alloys, and with the effects of these operations on the mechanical properties of the materials. The subjects covered are annealing, solution treating, stress relieving, stress equalizing, age hardening, hot working, cold working, combinations of working and heat treating (often referred to as thermomechanical treating), and properties of the materials at various temperatures. The equipment and procedures used in working the materials are discussed, along with the common problems that may be encountered and the precautions and corrective measures that are available.

Director
Technology Utilization Office

Acknowledgment

The information on which this report is based was obtained from technical articles, reports on Government-sponsored research programs, and manufacturers' literature. Additional data were acquired by personal contacts with technical personnel.

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Contents

CHAPTER 1

Principles

Because of its face-centered cubic crystallographic structure, nickel has a high degree of ductility and toughness over a wide range of temperatures from elevated to cryogenic and, for this reason, the metal responds readily to all commercial hot-working and cold-forming operations. In addition, the metal can be cast, welded, brazed, and soldered. Furthermore, nickel work hardens, so that its mechanical properties can be altered by mechanical treatment at room temperature. They also can be changed by hot working and by thermal treatment (refs. 1 and 2).

Another significant property of nickel is its ability to alloy readily with numerous other metals. As a result, it serves as the basic component of a great many alloys, important in military and aerospace applications and in the civilian economy. Additions of copper, molybdenum, and chromium, singly and in a number of combinations, enhance specific corrosion resistance. Chromium, tungsten, molybdenum, columbium, and cobalt, in various combinations, impart heat resistance and strength at elevated temperatures. As in other metallic systems, however, the introduction of alloying elements, especially those that have a strengthening effect, tends to reduce ductility and decrease workability, correspondingly.

The addition of titanium, aluminum, columbium, silicon, magnesium, and beryllium makes the material age hardenable. In nickel-base alloys, the age-hardening process is usually capable of effecting significant increases in room-temperature mechanical properties as well as producing tremendous increases in resistance to creep and rupture at elevated temperatures (ref. 2).

The purpose of this report is to provide information of a practical nature that is applicable to current practices and problems associated with the thermal and mechanical treatment of nickel and nickel alloys, attention being directed particularly to the influence of such treatments on mechanical properties.

The compositions of the nickel-base alloys that are discussed are given in table 1. These rather complex alloys were designed primarily to combine strength with corrosion resistance or to have superior resistance to creep and rupture at elevated temperatures. However, several of the alloys have, in addition, been useful at cryogenic temperatures. All of them comprise a group that is of particular interest in missile and aerospace applications.

Some of the alloys are intended to be used as castings and, thus, are not subjected to mechanical treatment. However, they respond to thermal treatment and are normally given a heat treatment before being placed in service.

Other alloys in the group are intended primarily to be used in the wrought condition, usually as machined forgings or components formed from sheet metal or bar stock. The mechanical properties of these alloys can be altered by hot working as well as by cold deformation. Likewise, these alloys respond to a variety of heat treatments with corresponding changes in their mechanical properties.

Neither experimental alloys nor alloys still in the developmental stage are included in the group. The alloys discussed are commercially established materials with an accumulated background of usage. In fact, in military and aerospace applications, they are among the prominent materials in present-day use. On the other hand, the list is not meant to be complete; rather, it is intended to be thoroughly representative of the nickel-base alloys now being used.

In the chapters that follow, consideration is given to the normal thermal and mechanical treatments for nickel and nickel-base alloys. These include an-

TABLE 1–*Selected Nickel-Base Alloys*

Name	Originator	Form	Nominal Composition, percent					
			C	Cr	Co	Mo	W	Ni
Wrought Nickel	The International Nickel Co., Inc.	Wrought	0.06	–	–	–	–	99.5[c]
Monel K-500	The International Nickel Co., Inc.	Wrought	0.15	–	–	–	–	(65)
Inconel X-750	The International Nickel Co., Inc.	Wrought	0.04	15.0	–	–	–	(73)
Alloy 718	The International Nickel Co., Inc.	Wrought	0.04	18.6	–	3.1	–	(52.5)
René 41	General Electric Co.	Wrought	0.12[a]	19.0	11.0	9.75	–	(51)
		Cast	0.09	19.0	11.0	9.75	–	(50.5)
Waspaloy	*Pratt & Whitney Aircraft	Wrought or cast	0.10[a]	19.5	13.5	4.25	–	(56)
Udimet 500	Special Metals Corp.	Wrought	0.15[a]	17.5	16.5	4.0	–	(52)
		Cast	0.10[a]	19.0	18.0	4.0	–	–
Hastelloy C	Union Carbide Corp.	Wrought	0.08[a]	15.5	2.5[a]	16.0	3.75	(56)
		Cast	0.12[a]	16.5	2.5[a]	17.0	4.5	(53.5)
Alloy 713C	The International Nickel Co., Inc.	Cast	0.12	12.5	–	4.20	–	(74)
Alloy 713LC			0.05	12.0	–	4.50	–	(75)
IN-100	The International Nickel Co., Inc.	Cast	0.18	10.0	15.0	3.0	–	(60)
MAR-M 200	Martin Marietta Corp.	Cast	0.15	9.0	10.0	–	12	(60)
TAZ-8	NASA, Lewis Research Center	Cast or wrought	0.125	6.0	–	4.0	4.0	(68)
TAZ-8A			0.125	6.0	–	4.0	4.0	(68)

(a) Maximum
(b) Low
(c) Minimum
Compositions in parenthesis are estimated balances
(d) Given as Cb + Ta
*Division of United Aircraft Corp.

TABLE 2–*Selected Nickel-Base Alloys*–Concluded

Cu	Fe	Al	Ti	Cb	V	B	Zr	Ta	Reference
0.05	0.15	–	–	–	–	–	–	–	3
29.5	1.00	2.80	0.50	–	–	–	–	–	3
0.05	6.75	0.80	2.50	0.85[d]	–	–	–	–	3
0.10[a]	18.5	0.40	0.90	5.0[d]	–	0.004	–	–	3,13
–	5.0[a]	1.50	3.15	–	–	0.006	–	–	7,8
–	5.0[a]	1.65	3.15	–	–	0.003[a]	–	–	7,8
0.50[a]	2.0[a]	1.30	2.90	–	–	0.006	0.07	–	9
0.15[a]	4.0[a]	2.90	2.90	–	–	0.010[a]	–	–	5,6
0.10[a]	2.0[a]	3.00	3.00	–	–	0.010[a]	–	–	
–	6.0	–	–	–	0.35[a]	–	–	–	4
–	6.0	–	–	–	0.30	–	–	–	
–	(b)	6.10	0.80	2.20	–	0.012	0.10	–	11
–	(b)	5.90	0.60	2.00	–	0.010	0.10	–	12
–	(b)	5.5	5.0	–	1.0	0.015	0.05	–	10
–	(b)	5.0	2.0	1.0	–	0.01	0.05	–	101
–	(b)	6.0	–	–	2.5	–	1.0	8.0	85
–	(b)	6.0	–	2.5	–	0.004	1.0	8.0	92

nealing, solution treatment, stress relieving, stress equalizing, age hardening, hot working, cold working, and combinations of thermal and mechanical treatments. Definitions of terms are supplied and data are provided showing the effect of the treatment on mechanical properties. In addition, properties of the materials at cryogenic temperatures are presented. Equipment and procedures are described as are the problems commonly encountered and the precautions and corrective measures that have been developed.

Many of the treatments have innumerable variants serving special purposes. Most of these are not discussed because of their limited applicability. Others are taken up to illustrate points of particular importance. In general, the practices discussed are those that have become standard for general usage, or are variants recommended by the producers to meet critical needs identified by leading fabricators or developed by the fabricators themselves.

Finally, the data for mechanical properties that are given in the report are intended to be illustrative or typical. In no case are they to be construed as minimum guaranteed or normal values. In many cases they are averages, but with an unknown statistical significance.

ANNEALING AND SOLUTION TREATING

Purpose Served

Annealing nickel and nickel-base alloys consists of heating at a suitable high temperature for a definite period of time and then cooling slowly or rapidly to soften the metal and increase its ductility (ref. 14). Annealing is commonly carried out on wrought material that has been hardened by cold forming such as deep drawing, stretch forming, or severe bending. In this case, softening and restoration of ductility are brought about by recrystallization of the metal's grain structure that occurs during annealing. The time and temperature employed in annealing encourage recrystallization. Nickel-alloy castings are often annealed also, but because they are not worked beforehand, they do not recrystallize during annealing.

The reasons for annealing nickel and nickel-base alloys are:

1. To increase the ductility and reduce the hardness of wrought alloys, improving their formability. Wrought nickel alloys work harden considerably during cold forming. Consequently, if the forming operation is severe, it is performed in steps and the metal is annealed between each step. The latter operation is often termed intermediate annealing or process annealing.
2. To facilitate machining, in cases where machining is not feasible because of the extreme hardness of the metal. However, it is often undesirable to anneal nickel-base alloys before machining because they tend to be "gummy" in the soft condition. Therefore, in numerous cases, machining hardened material is preferable.
3. To remove or reduce stresses built up in the material during casting, forming, or welding operations. (The subject of stress relieving is taken up in some detail in a later section.)
4. To soften the material for welding. Annealing reduces the tendency of weldments to crack from the combination of residual stresses produced by previous cold working with those due to the heating and cooling that accompanies welding.
5. To develop desired final mechanical properties. In addition to increasing ductility and decreasing hardness, annealing promotes toughness in nickel and nickel-base alloys.
6. To change the grain size of the metal. Annealing can either increase or decrease grain size, depending on the degree and the temperature of prior working, and the temperature and time of annealing.
7. To dissolve second phases in the matrix to improve corrosion resistance or prepare the alloy for subsequent age hardening. In these cases the annealing is known as solution treating or solution annealing (ref. 15). To be effective, however, the cooling rate must be rapid enough to retain the second phase in solution as the metal cools. Water or oil quenching is common; frequently, cooling in an air blast is sufficient, depending on the section and composition of the alloy.

Most nickel-base alloys used in missile and aerospace applications, such as those shown in table 1, are age hardenable. For these alloys, annealing and solution treating are essentially the same since they are seldom annealed without being cooled rapidly from the annealing temperature. With such alloys, the operation is often called "annealing" when the purpose is to soften the metal, or "solution treating" when the objective is to take second phases into solution.

Comments on Some Aspects of Annealing and Solution Treating

Grain Size and Solution Temperature

As indicated in the previous section, the grain size of the metal can be changed during annealing. In fact, nickel-base alloys, in common with most other metals and alloys, are susceptible to the growth of an exceedingly large grain structure if they are solution treated after small amounts of cold or hot work. To avoid growth of abnormally large grains on annealing, a critical amount of cold work (usually 1% to 6%, depending on the alloy) or hot work (generally about 10%) prior to solution treating must be exceeded in all areas of the part (ref. 15).

With larger amounts of prior cold work the grain size can be refined; however, the annealing temperature and time should not greatly exceed the minimum conditions for recrystallization. Again, the effectiveness of the grain-refining operation increases, the larger the prior grain size and the greater the amount of cold reduction before annealing.

Grain growth can be achieved by using annealing temperatures and times in excess of the minimum needed for recrystallization; the influence of temperature is far greater than that of time at temperature. Hence, increases in grain size are generally made by increasing the temperature, rather than the time, at which the work is annealed or solution treated. However, even when the metal is heavily cold worked before annealing, there are limitations on the annealing temperature if excessively large grains are to be avoided. Annealing at a temperature somewhat above that at which all second phases are completely dissolved stimulates rapid grain growth. As an example, it has been reported that annealing at 1700° to 1800° F adequately dissolves precipitated phases in Inconel 718. However, when this alloy is annealed at temperatures above 1900° F, excessive grain growth can be expected (ref. 16).

An abnormally large-grain structure (i.e., ASTM No. 1, 0, or greater) should be avoided in nickel and its alloys because it reduces creep and fatigue properties and interferes with cold forming. It also reduces uniform elongation and encourages the development of "orange peel" on surfaces where cold forming causes a considerable amount of bending or stretching. Cold forming is favored by a uniform fine-grained structure, provided the solution-treating temperature is high enough to completely dissolve age-hardening constituents and achieve the soft condition. If the minimum solution treatment does not make the metal soft enough for a particular forming operation, the temperature must be raised to soften the metal but a corresponding increase in grain size must also be accepted (ref. 15).

Often a solution temperature at the lower end of the range is used to achieve optimum low-temperature and short-time elevated-temperature properties in material to be subsequently aged. The fine-grain structure produced during the solution treatment is considered to be a major factor in the development of these properties in the aged material. By the same token, a higher solution temperature, which causes a coarser grain structure, is often specified for optimum creep and rupture properties.

Cooling from the Solution Temperature

For maximum softness as annealed, and for optimum aging response, most age-hardenable nickel-base alloys should be cooled rapidly from the heating temperature. Delay in cooling, or a cooling rate that is too low, may cause partial precipitation of the aging phase. As a result, the material may not be sufficiently soft on reaching room temperature and may not respond to a subsequent aging treatment. Alloys such as René 41 and Waspaloy, which contain large amounts of hardening elements, are particularly sensitive to delayed cooling and to inadequate cooling rates.

These alloys cannot be uniformly softened in thick sections even by drastic water quenching from the solution temperature. Partial precipitation of the hardening phase occurs in the interior of the material. To obtain more uniform hardness after solution treatment and more uniform response to aging, heavy sections of these alloys are often air-cooled from the solution temperature in spite of the partial aging that occurs from slower cooling. Sometimes, to obtain uniform softness in parts composed of thick and thin sections, the thick sections are quenched more drastically than the thin sections. For example, the thick sections may be immersion quenched with water, while the thin sections are spray quenched (ref. 15).

Effect of Intermediate Annealing

In addition to softening the material and restoring its ductility so as to facilitate further cold forming, intermediate or process annealing can sometimes have a marked effect on the response to subsequent heat treatment. René 41 and Waspaloy are reported to be quite sensitive to such effects.

Parts formed from Waspaloy and René 41 sheet have cracked after being solution treated at 1975° F for 1/2 hr, air-cooled, and then aged at 1400° F for 16 hours. The cracking was attributed to a carbide network in the grain boundaries, which drastically reduced the ductility of the material. The cause was traced to a process anneal at 2150° F. At this temperature the M_6C carbides present in random dispersion throughout the grains were dissolved, making the carbon available for the precipitation of $M_{23}C_6$ carbides at the grain boundaries on subsequent exposure of the material to temperatures in the range of 1400° to 1600° F. By keeping the annealing temperature below 2000° F, the M_6C carbides were not dissolved and the ductility of the aged metal was improved (ref. 15).

Equipment and Procedures

Furnaces

In this section the principal characteristics of the equipment used to anneal nickel and nickel-base alloys are discussed. Details on the types of furnaces and their construction, and the auxiliary equipment involved, are not presented. There are numerous manufacturers of industrial heating equipment, and annealing equipment has countless variations; while much of the equipment can be considered as standard, a great deal of it is specially designed or has special features to suit particular situations. Accordingly, annealing equipment is discussed here only in general terms, but with particular attention directed toward any special requirements imposed because the equipment is handling high-nickel material.

Wrought nickel can be annealed by all standard muffle-furnace, box, and salt-bath processes. Roller-

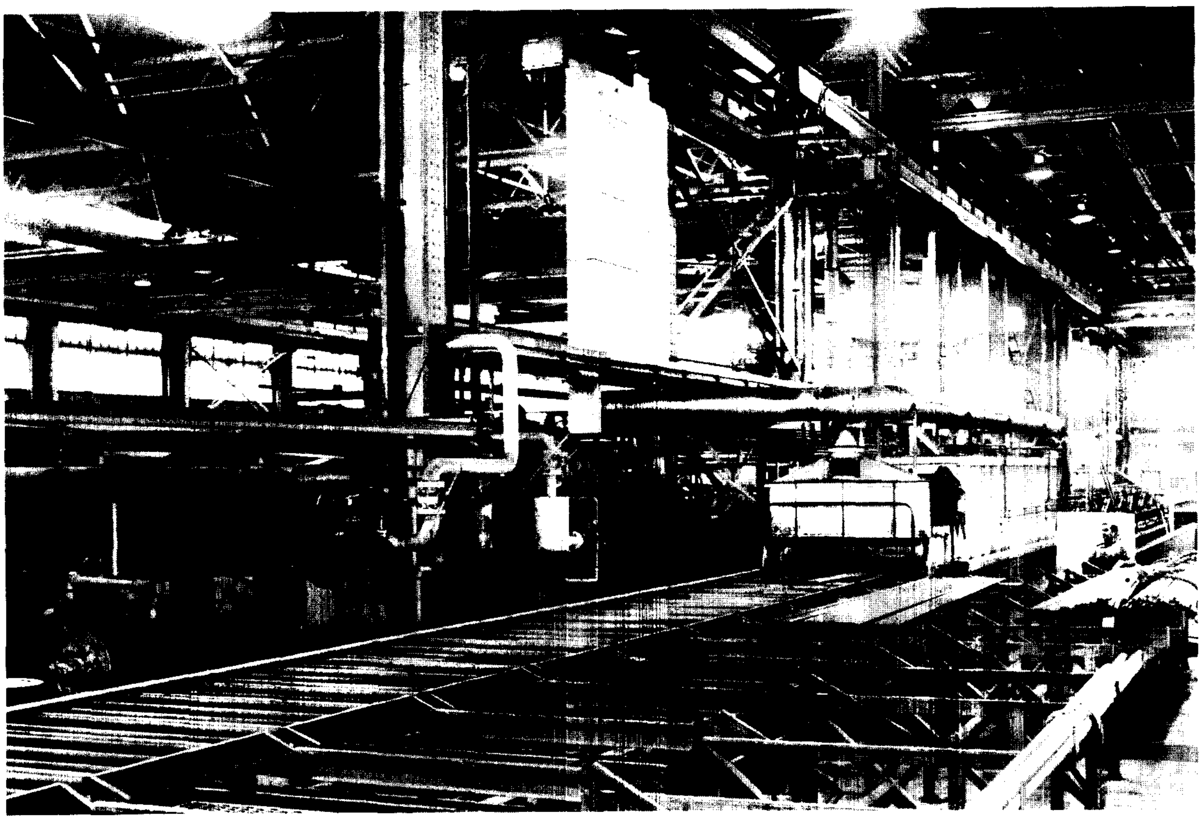

FIGURE 1.–Continuous roller-hearth furnace used for annealing nickel-base alloys. (Courtesy International Nickel Co.)

hearth (fig. 1) and belt-conveyor open-muffle furnaces are common. When open-muffle furnaces are used, the material to be annealed is protected from oxidation by a prepared atmosphere pumped into the furnace; or, in the case of fuel-heated furnaces, by the products of combustion. Temperatures are comparatively high, and time periods are quite short. Because the annealing periods are short, temperature control is critical (ref. 17).

Nickel is box annealed in bell- or pot-type equipment. In the bell-type box, the work is loaded on the flat base and the bell is lowered over the charge. In the case of pot furnaces, the charge is lowered into the pot and then the lid is put in place. Heat-resistant nickel-chromium and nickel-chromium-iron alloys are recommended as box materials not only because they can withstand the service but also because they do not shed loose scale or rust on the charge. Fire clay and sand are common sealants for box furnaces, which are usually heated electrically, but oil and gas are also used. A fuel-fired furnace should be designed to avoid the direct impingement of flames on the surface of the container, and the container should have an inlet to admit gas. Reducing gas is introduced as soon as the box is removed from the furnace or when the burners are shut off, the flow of gas being regulated to maintain a positive pressure. Temperatures for box annealing are relatively low and time periods are fairly long and, thus, temperature control is less critical than for open-muffle annealing (ref. 14).

Small nickel parts may be salt-bath annealed. The bath is usually a mixture of molten inorganic chlorides or chlorides and carbonates held in steel pots, heat-resisting nickel-chromium-iron alloy pots, or ceramic containers, depending on the temperature. When placed in the bath the work heats rapidly. Annealing cycles are short and, after annealing, the work is water quenched to remove entrained particles of salt (ref. 14). Examples of salt-bath mixtures used for nickel alloys are shown in table 2 (ref. 18).

In general, nickel-base alloys can be annealed in muffle-type equipment and in salt baths. Those alloys that require solution treating, i.e., rapid cooling from the annealing temperature, must be annealed in furnaces provided with suitable cooling chambers or quenching equipment. This applies to age-hardenable alloys such as those listed in table 1. Thus, box annealing is usually unsuitable for these alloys.

TABLE 2–*Neutral Salt-Bath Mixtures for Age Hardening and Annealing Nickel and Nickel-Base Alloys*

Mixture	Percent by Weight	Melting Point, °F (approx)	Working Range, °F
Sodium nitrite	40-50	290	325-1200[a]
Sodium or potassium nitrate	60-50	290	325-1200[a]
Sodium nitrate	40-50	440	500-1200[a]
Potassium nitrate	60-50	440	500-1200[a]
Sodium nitrate	96 min	700	750-1200[a]
Potassium chloride	20-30	1100	1250-1700
Sodium chloride	15-25	1100	1250-1700
Barium chloride	50-60	1100	1250-1700
Potassium chloride	45-55	1250	1350-1650
Sodium chloride	55-45	1250	1350-1650
Sodium chloride	20-30	1300	1400-1700
Barium chloride	80-70	1300	1400-1700
Sodium chloride	10-20	1400	1500-2000
Barium chloride	90-80	1400	1500-2000
Barium chloride	85	1550	1650-1850
Barium fluoride	15	1550	1650-1850

(a) Do not operate this bath above 1100° F in a fuel-heated pot

Protective Atmospheres

In the annealing of nickel and nickel alloys it is generally advisable to avoid a strongly oxidizing atmosphere. It is even more important that the atmosphere be suitably sulfur-free, because a variety of sulfide inclusions can be formed in nickel and nickel alloys (ref. 19). Frequently, these inclusions take the form of grain-boundary networks that embrittle the material. The nature of such a grain-boundary attack as it occurred in heating wrought nickel in a high-sulfur furnace atmosphere is shown in figure 2 (ref. 20). With nickel-base alloys the character of the attack is generally quite similar to that depicted in the figure; i.e., the resulting intergranular nickel-sulfide network has been darkened by the etchant.

Wrought nickel will remain bright when heated and cooled in a reducing atmosphere. Nickel-base alloys

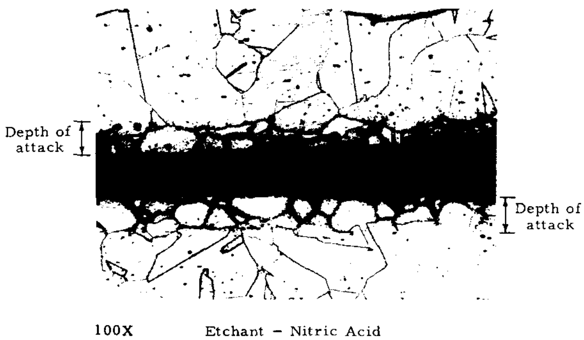

FIGURE 2.–Transverse sections showing intergranular attack of wrought nickel when heated to high temperature in a sulfur-containing furnace atmosphere.

containing chromium, titanium, and aluminum, however, will form a thin oxide film unless they are annealed in a vacuum or in a very pure, very dry inert gas (ref. 17). In many instances, a thin oxide film can be tolerated. This is often the case in applications involving heat resistance. When not acceptable, the oxide can be removed by pickling or by salt-bath descaling and pickling.

A commonly used and economical atmosphere, which can be provided in fuel-fired muffle-type furnaces, is that produced by controlling the ratio of fuel to air that is fed to the burners so that there will be a slight excess of fuel. It is desirable that the products of combustion contain at least 4% carbon monoxide plus hydrogen and no more than 0.05% uncombined oxygen. An atmosphere that fluctuates between reducing (the presence of carbon monoxide and hydrogen) and oxidizing (excess air) is to be avoided. This tends to cause intergranular attack and embrittlement of the metal even when no sulfur is present. Nickel and Monel are more readily affected than nickel-base alloys containing chromium (ref. 17).

Prepared endothermic atmospheres produced by reacting fuel gas with air in the presence of a catalyst are not recommended because they tend to be carburiziñg. Likewise, the endothermic mixture of nitrogen and hydrogen formed by dissociating ammonia is not to be used unless the dissociation is complete, because of the chance that it will nitride the work (ref. 15).

Prepared atmospheres suitable for nickel and nickel-base alloys include dry hydrogen, dry nitrogen, completely dissociated ammonia, and dry argon. The general practice is to introduce prepared atmospheres into indirectly heated annealing equipment rather than into a direct-fired furnace.

Dry hydrogen with a dew point of −60° F or lower is used for bright or semibright annealing of nickel and nickel-base alloys. However, alloys containing appreciable aluminum or titanium tend to show a very light tarnish. Hydrogen is not recommended for annealing or solution treating of alloys that contain boron because of the likelihood of "deboronizing" through the formation of boron hydrides (ref. 15).

When nickel-base alloys containing strong oxide formers such as aluminum and titanium, with or without boron, must be bright annealed, a vacuum or an inert gas such as argon is required. When argon is used it must be pure and must have a dew point of −60° F or lower. Some atmospheres used in the annealing and solution treatment of nickel and nickel-base alloys are shown in table 3 (ref. 17).

Fuels for Heating

Nickel and nickel-base alloys may be heated by electricity or by the combustion of a number of fuels. When the latter are used they must be carefully selected. Coal and coke are generally unsatisfactory because their combustion is too inflexible to lend itself to sufficiently close temperature control. In addition, the products of combustion of these fuels usually contain an unacceptably large amount of sulfur compounds. Oil is satisfactory provided it has a low-sulfur content. Burners that use low-pressure air, supplied through the burner, are preferred over high-pressure burners or those of the steam-injector type because they work especially well with automatic temperature-control equipment.

Gas, whether natural or manufactured, is usually the preferred fuel. Good heating can be obtained readily with gas because of the ease with which the gas-air mixture can be controlled and the supply of gas regulated. Gaseous fuels require only a small combustion space and automatic control of temperature and atmsophere is readily accomplished. Generally speaking, natural gas is the best because it can be obtained, in many localities, essentially free from sulfur compounds. Manufactured gases are produced from coal or oil, which often contain substantial amounts of sulfur. These gases should not be used unless the sulfur compounds are effectively removed during gas manufacture. The generally accepted statutory limit

TABLE 3—*Some Atmospheres Used for Annealing and Solution Treating Nickel and Nickel-Base Alloys*

Description	Air-to-Gas Ratio(a)	Composition, percent by volume						Dew Point, °F (approx)	Characteristics of Atmosphere(c)
		H_2	CO	CO_2	CH_4	O_2	N_2		
Completely burned fuel, lean atmosphere	10:1	0.5	0.5	10.0	0.0	0.0	89.0	Saturated(d)	Noncombustible, very slightly reducing
Partially burned fuel, medium rich atmosphere	6:1	15.0	10.0	5.0	1.0	0.0	69.0	Saturated(d)	Combustible, reducing
Dissociated ammonia (complete dissociation)	No air	75.0	0.0	0.0	0.0	0.0	25.0	-70 to -100	Combustible, strongly reducing
Dissociated ammonia (partially burned)	1.25:1(e)	15.0	0.0	0.0	0.0	0.0	85.0	Saturated(d)	Combustible, strongly reducing
Dissociated ammonia (completely burned)	1.80:1(e)	1.0	0.0	0.0	0.0	0.0	99.0	Saturated(d)	Noncombustible and inert
Electrolytic hydrogen, dried by alumina + molecular sieves	–	100.0	0.0	0.0	0.0	0.0	0.0	-70 to -100	Combustible, strongly reducing

(a) These ratios represent natural gas containing nearly 100% methane. For high-hydrogen manufactured gas (550 Btu) these ratios are about one-half the values given above. For manufactured gas with lower hydrogen and high CO content (450 Btu), ratios are about 40% of values. For propane, values are approximately twice the values given in the table; for butane, multiply by 3.
(b) Volumes represent high-methane natural gas. Double these values to approximate various types of manufactured gas. For propane, requirements are one-half of values given, while butane requires one-third.
(c) Particularly with respect to nickel and high-nickel alloys.
(d) Dew points of these atmospheres, when cooled by tap-water heat exchangers, will be about 10° to 13° F above the tap-water temperature. Dew point may be reduced to about 40° F by refrigeration equipment; to minus 70° F or even less by activated absorption equipment.
(e) Ratio—air to dissociated ammonia.

of 30 grains of sulfur/100 cu ft of gas is a little too much for heating nickel alloys; a sulfur content of 10 to 15 grains is far more acceptable. Two other good fuels for heating nickel and nickel-base alloys are propane and butane (ref. 17).

Notes on Procedures

Nickel and nickel alloys are subject to embrittlement when heated in the presence of sulfur and sulfur compounds. Thus, fuels for heating these materials must have a low sulfur content. Many lubricants, oils, and greases contain sulfur or lead that will have the same embrittling effect so must be removed before heat treating. Similarly, paints and other adhering substances may contain lead, sulfur, and other harmful ingredients and should be removed beforehand. In fact, most foreign material tends to burn into the metal at high temperatures. A good cleaner is hot trisodium phosphate solution; mineral oils and greases may be removed with carbon tetrachloride and similar solvents as well as with vapor-degreasing machines (ref. 17).

Annealing and solution-treating temperatures are usually well above the upper limit of usefulness of the material, even of alloys specifically designed for strength at high temperatures. At such temperatures the metal is weak, and tends to sag and distort under its own weight and also as a reaction to the relief of internal stresses. To counteract these effects, parts and assemblies are often held in fixtures during heat treatment. Some types of fixtures merely support the work against the action of its own weight; others apply restraint to minimize distortion. Long, reasonably symmetrical pieces are frequently hung vertically in a fixture. Flat parts, or assemblies with large flat surfaces, are often placed in the furnace on a flat tray or plate. Irregularly shaped parts may be supported on a bed of clean sand or a ceramic casting formed to the shape of the part, or on suitably shaped refractory bricks. Restraint fixtures usually have grooves, lugs, or clamps to hold the part to a given shape. Generally, for alloys to be quenched from the solution temperature, a minimum of fixturing is used, warpage and distortion being corrected during subsequent aging (ref. 15).

Control of temperature is always important, especially in continuous annealing and salt-bath annealing where temperatures are high and time periods are short. Good temperature indicating, controlling, and recording equipment is available. Thermocouples should be checked daily and replaced regularly. It is usually advisable to supplement the furnace-control pyrometer with indicating or recording equipment that measures, or closely approximates, the temperature of the work. This is the important temperature and it is frequently different from that of the furnace.

Cooling, the final operation in a heat-treating cycle, may be carried out in several ways depending on requirements. For example, if retention of second phases in solid solution in the matrix is not important, the cooling rate is generally not critical and any convenient method, such as furnace cooling or air cooling, is satisfactory. However, considerations of economics and time generally favor rapid cooling methods. Some alloys call for rapid air cooling or air quenching; this means cooling with a fan or an air blast. For water or oil quenching, the equipment is similar to that used for heat-treating hardenable steels. A generous amount of the quenching liquid should be available and it should be circulated vigorously in the tank. Intermediate cooling rates can be obtained by using a water spray or a combination of water and compressed air.

STRESS RELIEVING

Purpose

A stress-relieving heat treatment should remove or reduce internal stresses in a metallic material or an assembly of metallic components for these reasons:

1. To promote dimensional stability and minimize warping and distortion from the action of, or the release of, internal stresses during such subsequent fabricating steps as machining and welding, or in service.

2. To increase the capability of the material to withstand the stresses imposed during subsequent forming operations without cracking or fracturing. Eliminating internal stresses removes the danger that they may act in concert with the applied stresses to cause failure.

3. To avoid cracking during welding. Previously introduced internal stresses are not available to reinforce stresses developed during welding.

4. To avoid cracking during storage or service, especially where there is danger of stress-corrosion cracking.

Procedures and Effects

According to the Metals Handbook stress relieving consists of heating to a suitable temperature, holding long enough to reduce residual stresses, and then cooling slowly enough to minimize the development of new residual stresses (ref. 21). Clearly, it is feasible to stress relieve by this procedure only simple materials that do not undergo harmful changes on slow cooling from elevated temperatures. Wrought nickel can be, and is, stress relieved in this manner.

Hot-worked or cold-worked nickel is generally stress relieved in the range of 900° to 1300° F, the time at temperature varying from 3 hr or so at the lower temperatures to 1/2 hr or less at the upper end of the temperature range (refs. 14 and 17). Time and temperature are often quite critical and usually must be determined by experimentation. The amount of prior working and the degree of stress relief desired are controlling considerations. In general, the principle is to select a temperature and time to bring about stress relief without producing a recrystallized grain structure and without altering the mechanical properties to any great extent. Stress relief takes place principally by means of creep.

More complex materials in which carbides, intermetallic compounds, and other phases precipitate during slow cooling from elevated temperatures are generally not given a true stress-relieving heat treatment. Age-hardenable alloys, for example, are in this category. In such alloys, the precipitation of second phases during the slow cooling period may either impair corrosion resistance, reduce response when subsequently aged, or promote brittleness.

Because of these effects, the stress relieving of age-hardenable nickel-base and similar alloys is avoided where possible. Where relief of stress is absolutely necessary, for example, following the welding of a highly restrained joint, one of two recourses is usually resorted to. If the assembly is relatively simple and symmetrical so that it can be heated to high temperatures and rapidly cooled without cracking or undue distortion, it may be given a full annealing or solution treatment instead of a true stress-relieving heat treatment. On the other hand, if the configuration does not permit annealing, aging can be used.

Equipment

Stress relieving can be carried out in continuous heat-treating or batch-type equipment. Continuous open-muffle furnaces are best adapted to the higher temperature, shorter time cycles and to the lighter sections. For cycles involving the longer times and lower temperatures, box and car furnaces are employed.

When a solution treatment is used as a substitute for a true stress-relieving treatment, the equipment must have facilities for quenching or other rapid cooling. Because high temperatures are involved, a suitable furnace atmosphere is required. Standard solution-treating equipment, such as that described earlier, is generally used.

Fixtures of the support or restraint type may or may not be employed in stress relieving depending on a number of circumstances. When relatively low temperatures are used, for example, the work will be strong enough to make supporting fixtures unnecessary. Likewise, if dimensional changes can be tolerated because other forming operations are to follow, restraint fixtures may not be necessary.

STRESS EQUALIZING

Stress equalizing is a term sometimes used for a low-temperature heat treatment that brings about partial "recovery" in cold-worked nickel and nickel alloys and, in so doing, tends to balance and readjust residual stresses. This recovery, before any microstructural changes, produces an increase in proportional limit, a slight increase in hardness and tensile strength, no significant change in ductility, and a return of electric conductivity toward the value characteristic of the material in the annealed condition (ref. 17).

Stress-equalizing temperatures usually range from 500° to 900° F. The temperature required depends on the amount of cold work as well as on the composition of the alloy. Heavily cold-worked materials reach a higher energy level and, therefore, require a lower temperature (ref. 17).

Stress equalizing is usually applied to coil springs, wire forms, and flat-spring stampings. If coil springs are to be given a cold "set," the stress-equalizing treatment should precede the setting, which involves stressing the material beyond the elastic limit. Any cold-working stresses set up by this operation are in such a direction as to be beneficial, rather than harmful, to the material. If the stress-equalizing treatment were carried out after cold setting, part of the beneficial cold-working stresses would be removed (refs. 14 and 17).

Stress-equalizing treatments are not usual with alloys that are to be age hardened. When the aging treatment is at relatively low temperatures the effects of stress equalizing occur to a considerable extent, the aging operation acting as a sort of substitute for a stress-equalizing treatment. On the other hand, aging at comparatively high temperatures would tend to destroy many of the effects of stress equalizing. The most frequent use of stress-equalizing on age-hardenable nickel alloys is in cases where the alloy has been cold-worked after age hardening.

AGE HARDENING

Many nickel-base alloys develop high strength after heating at intermediate temperatures, usually in the range of 800° to 1600° F, when they have been solution-treated. This is known as age hardening. The increase in strength results from microstructural changes that take place within the metal. The microstructural change is the precipitation of one or more dispersed phases throughout the matrix. The precipitation actually takes place from the supersaturated matrix that is developed by the prior solution treatment and retained by rapid cooling from the solution-treating temperature (ref. 18).

For most age-hardenable nickel-base alloys, numerous sets of aging conditions (i.e., aging times and temperatures) are available. The choice depends primarily on the final mechanical properties desired,

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the number and characteristics of the available precipitating phases, and the degree to which the material may have been cold worked after the solution treatment. In addition, the aging conditions may be influenced by the prior solution-treating temperature.

Although maximum strength may not be required for a certain application, for example, it may be desirable to develop higher strength than is available in the material as solution treated, but greater ductility than is possessed by the fully aged alloy. In this case, the aging conditions that give maximum strength would not be used. Instead, the aging temperature or the aging time might be decreased to "underage" the material and effect a substantial increase in strength without a drastic reduction in ductility (ref. 18). On the other hand, desirable intermediate properties are sometimes achieved by "overaging" the material, that is, by using higher temperatures or longer times than optimum.

For a number of nickel-base alloys, especially those in which more than one precipitating phase is available, a double aging treatment has been developed. A treatment in the vicinity of 1550° F precipitates $M_{23}C_6$ carbides and $Ni_3(Al, Ti)$, while reheating in the range of 1300° to 1400° F results in optimizing the particle size of the $Ni_3(Al, Ti)$ precipitate and increasing the amount of this phase that is precipitated. By this means optimum creep and rupture properties are often obtained. However, the $M_{23}C_6$ carbides tend to precipitate along grain boundaries where they can cause a marked decrease in ductility. Thus, when the ultimate in high-temperature rupture and creep strength is not required, but more than minimum ductility is needed, the 1550° F aging treatment is omitted. It must be remembered, however, that subsequent service in the temperature range of 1500° to 1600° F or so will likewise cause intergranular precipitation of the $M_{23}C_6$ carbides with an attendant loss of ductility. As a consequence, the grain in ductility is realized for service at room temperature and at elevated temperatures below those at which $M_{23}C_6$ carbides precipitate (ref. 15). In a double aging sequence, the first treatment carried out at the higher temperature is sometimes called the high-temperature aging treatment, the intermediate aging treatment, or the stabilizing treatment.

By increasing the internal energy of the material and the density of precipitation sites, cold working accelerates aging. Thus, when cold-worked age-hardenable alloys are heated at aging temperatures that are optimum for the solution-treated material, the precipitates tend to appear early and overaging is likely to occur. Material, solution treated and then cold-worked, is normally aged at lower temperatures and shorter times than the same material not cold-worked. Thus, heavily cold-worked Monel K-500 may be aged at 1000° F for 6 to 10 hr, while the unworked alloy is aged at 1100° F for 16 hr (ref. 18).

The prior solution-treating temperature may be changed for any one of several reasons as indicated in the foregoing section. To achieve the desired type of properties to the optimum degree often requires development of aging conditions specifically tailored to the solution treatment. For example, to obtain maximum stress-rupture properties in René 41 bar stock, a solution treatment at 2150° F followed by aging 4 hr at 1650° F may be used; however, for maximum room-temperature tensile strength, the solution temperature is lowered to 1950° F and, at the same time, the aging conditions are changed to 16 hr at 1400° F (ref. 7).

In some nickel-base alloys the cooling rate from the age-hardening temperature is important. For Monel K-500, for example, achievement of optimum properties requires cooling at a controlled rate from the aging temperature, or cooling in steps spaced apart by a predetermined number of degrees of temperature (ref. 18).

Equipment and Procedures

Nickel alloys usually are hardened in box equipment, described in the section on annealing and solution treating, or in car-bottom furnaces, as illustrated in figure 3. Continuous furnaces are impractical because of the long times required for aging. Small parts are sometimes hardened in salt baths usually composed of mixtures of sodium, potassium, and barium chlorides. The compositions of some of the salt mixtures that are used are given in table 2.

Electric furnaces are often used because they offer temperature uniformity and control as well as freedom from contamination of the work. The next choice is gas-fired equipment, especially furnaces of the radiant-tube type.

An effective way to avoid oxidation during hardening is to use a vacuum or an ultrapure inert gas. This is feasible only for small parts and is not yet a general commercial method. Usually, no effort is made to control oxidation because, in most cases, the develop-

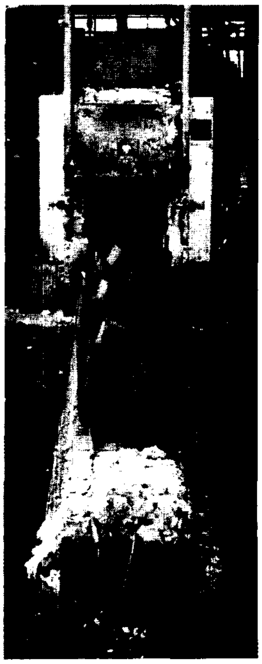

FIGURE 3.–Car-bottom furnace used for heating in age-hardening treatments. (Courtesy International Nickel Co.)

ment of an oxide film on the work is unimportant. Semibright hardening can be achieved, if deemed desirable, by using electrolytic hydrogen dried to a dew point of −70° F or less or by using completely cracked and dried ammonia. When bright or semibright hardening is not required, but extensive scaling is to be avoided, other atmospheres may be used such as nitrogen, cracked natural gas free of sulfur, cracked city gas, cracked hydrocarbons, or a generated gas. The use of sulfur-free gases is necessary to avoid embrittlement of the materials being aged (ref. 18).

All parts to be hardened should be free of dirt, lubricants, and other foreign matter.

EFFECTS AT CRYOGENIC TEMPERATURE

Metallic materials can be affected in several different ways when the temperature is reduced into the cryogenic range. For instance, the physical and mechanical properties of the material usually change as the temperature is decreased. In general, the tensile strength and the yield strength of nickel and nickel-base alloys increase moderately as the temperature is decreased below room temperature. In some cases the ductility is not greatly affected; in others it may decrease or increase.

Another cryogenic effect can be illustrated in the hardening of quenched- and tempered-alloy steels and semiaustenitic stainless steels. A step consisting of cooling to, and holding at, a cryogenic temperature is often inserted immediately after quenching and prior to tempering. It completes the martensite transformation and, by doing so, increases the strength of the material. Thus reduction of the material's temperature to cryogenic levels can be considered a thermal treatment. The majority of nickel alloys do not have low-temperature phase changes and transformations and, therefore, they are not benefited by such a cryogenic treatment.

Finally, metal forming has been carried out at cryogenic temperatures. In a sense, the operation can be classified as a type of thermomechanical treatment. Thus far it has been limited to AISI Type 301 and similar chromium-nickel stainless steels that can transform to martensite at low temperatures. When this type of steel is formed at cryogenic temperatures it increases more in strength than at room temperature. In addition, the steel displays a greater amount of uniform elongation when worked at cryogenic temperatures and, consequently, can be successfully formed in more severe operations at low temperatures than at room temperature. So far as is known, however, nickel and nickel-base alloys are not formed at cryogenic temperatures, and the supposition is that their deformation characteristics are not greatly altered by decreasing the temperature below ambient.

HOT WORKING

In this section, comments are made on hot and cold working. The objective is to point out factors and considerations peculiar to nickel and nickel-base alloys rather than to discuss hot- and cold-working operations in detail. Since this report is concerned primarily with the effect of the treatments on mechanical properties, a detailed discussion of the mechanics of the operations is beyond its scope.

Fuels

One of the most important factors in successfully hot-working nickel and nickel-base alloys is the selection of a satisfactory fuel to heat the material. As indicated in the section on annealing and solution treating, high-nickel alloys are susceptible to attack by sulfur during heating and, therefore, exposing them at high temperatures to sulfur-containing atmospheres or other sources of sulfur must be scrupulously avoided. The most common source of sulfur is the fuel and, therefore, care must be taken to use only low-sulfur fuels.

Low-sulfur natural gas is one of the best fuels. Manufactured gases low in sulfur are also satisfactory, as are butane and propane. Coal and coke are generally unsatisfactory not only because they are often high in sulfur but also because it is difficult to maintain the proper heating conditions when using these fuels (ref. 22).

Comments on Heating Conditions

The atmosphere surrounding the work while it is being heated to the hot-working temperature should not only be sulfur-free but also slightly reducing. A minimum content of 2% carbon monoxide is recommended. The reason for using a reducing atmosphere is to minimize scale formation; the scale built up on nickel-base alloys at high temperature is adherent and does not break away and slough off readily as the scale on plain-carbon and low-alloy steels does. In other words, nickel alloys are not "free scaling."

It is important to provide against other sources of sulfur in addition to the heating atmosphere. Slag and cinder on the hearth may contain sulfur; therefore, the metal is supported on rails or other suitable shapes. For the same reason, pickling salts, oil, grease, and marking paints are removed from the metal before heating (ref. 22).

The metal should be heated at the recommended temperature. Operating the furnace excessively hot to speed up production is ill advised, as is raising the heating temperature to compensate for loss in temperature on transferring the work to the hammer or other forming equipment. Both practices can bring about incipient melting of the metal. The stock should be turned frequently to aid uniform heating, and direct-flame impingement should be avoided.

The heating time should be no more than necessary to insure that the work has been thoroughly heated through to the proper temperature. For most alloys, no more than 1/2 hr/in. of thickness is necessary. Soaking is not recommended (ref. 22).

Notes on Working

Hot working should begin immediately after pulling the stock from the furnace. Even a short time lapse may allow surface temperature to drop 100° to 200° F (ref. 23).

In forging nickel-alloy ingots, light rapid blows are used until the cast structure is fairly well refined. After perhaps 40% reduction, heavier blows may be struck. In the initial stages of working, the general shape of the ingot is not changed substantially. For example, in making a round from a square ingot, the metal is first worked into the shape of an octagon, then rounded.

Hammer forging has some advantages over press forging. The metal can be worked more rapidly and the time the metal is in contact with the dies can be held to a minimum; thus, the opportunity for localized chilling is reduced.

Forging temperature and percent reduction during forging may have a profound effect on final mechanical properties, largely through the influence of these variables on grain size. Particularly important is the finishing temperature and the amount of reduction in the final pass, i.e., the lower this temperature and the greater the reduction, the finer will be the grain size (ref. 22).

It is important to work the metal within the recommended temperature range. Working too hot can cause burning and incipient melting, while working the metal too cold may cause it to crack and split. Age-hardenable alloys should be fast cooled after hot working. If slow cooled through the age-hardening range, they will partially harden. This will make machining difficult and also make it necessary to solution anneal to achieve proper response when subsequently aged.

Nickel-base alloys can be hot bent successfully as well as hot upset and impact extruded. These operations, however, are not carried out on ingots or other castings because of the probability of cracking. Ingots are generally worked at least 75% before undergoing other hot-forming operations (ref. 23).

Finally, when cracks or tears develop during hot working, they are removed because they will not heal during subsequent working. Rather, they will worsen. A frequently satisfactory way to remove cracks and tears is to grind them out while the metal is still hot instead of waiting until the metal is cold and either chipping or grinding. Some alloys tend to heat check if ground cold, especially when they are cast or only lightly worked.

COLD WORKING

Mill Products

Wrought nickel and most nickel-base alloys can be cold worked successfully and, consequently, are available from the mill in forms such as cold-rolled sheet and strip, and cold-drawn rod, tubing, and wire. Because nickel work hardens considerably and wrought nickel alloys often work harden to a great extent, substantially higher strength is usually available in a cold-worked form than in annealed material. On the other hand, the increase in yield strength and ultimate tensile strength resulting from cold working is obtained at a corresponding sacrifice in ductility and toughness.

In common with most other wrought metals and alloys, cold-rolled nickel and nickel-alloy sheet and strip are anisotropic in their mechanical properties. In particular, the ductility and toughness in the transverse direction are usually substantially less than in the rolling direction. The degree of anisotropy is strongly influenced by the rolling and process-annealing schedules.

There is also a tendency toward the development of a grain structure in sheet and strip that has a preferred orientation. In fact, a preferred grain orientation may persist even when the metal is fully annealed after being cold-rolled. These characteristics are shared with many other metallic materials. Sheet that has a grain structure showing pronounced preferred orientation does not deform in a uniform manner and shows the familiar "earing" phenomenon when cold-formed in certain ways, especially when deep-drawn. As is the case with anisotropy in mechanical properties, the degree of preferred orientation displayed in sheet material is strongly influenced by the frequency of intermediate anneals, the amount of reduction between anneals, the extent of cross rolling, and other factors.

Wrought nickel and some nickel-base alloys are available in a number of different tempers, that is, a number of different degrees of cold working. This is especially the case for sheet and wire. And, of course, the more heavily cold-worked material has higher strength and lower ductility and toughness.

In addition, some wrought age-hardenable nickel-base alloys can be obtained as cold-rolled or as cold-drawn and then aged. The usual forms are strip and wire. By cold working the material and then aging it, tensile strength in the vicinity of 200,000 psi and above can be reached. This combination of solution treating, cold working, and aging is the only type of thermomechanical treatment applied to nickel-base alloys in commercial practice.

Cold Forming

Most wrought nickel-base alloys can be cold formed to some degree. They tend to work harden rapidly and, therefore, require frequent intermediate anneals. For age-hardenable alloys, these anneals should actually be solution treatments. These alloys can be press formed or drawn, or spun if a sufficient number of process anneals is used.

Occasionally the material is placed in service as cold-formed. Age-hardenable alloys may sometimes be put in service as cold-formed and aged to take advantage of the high strength developed by this combination of treatments. Springs made of age-hardenable nickel-base alloys are usually processed in this manner.

On the other hand, in many cases, the amount of cold working and, hence, strengthening, is nonuniform and not well known. Rather than place any reliance on the strength obtained from this source, it is usually better practice to solution treat and age the finished part. Again, the strength imparted by cold working diminishes and tends to disappear when the material is used at elevated temperatures.

CHAPTER 2

Effects of Treatments on Mechanical Properties

The preceding discussion has directed attention to the variety of thermal and mechanical operations that may enter into the fabrication of nickel and nickel alloys. These all influence the mechanical properties of the materials to some extent, and a knowledge of the effects is of vital importance in service. These investigations show that determination of mechanical properties is undertaken with various objectives in mind, including these:

1. To provide nominal data for inclusion in brochures published by producers of the alloys
2. To provide minimum or guaranteed property data that are useful for designers and engineers
3. To determine variation in properties between different heats of metal from one producer
4. To compare properties obtained on the same type of alloy from several producers
5. To determine whether commercial material has the desired properties for specific applications from data obtained by contractors and other prospective consumers of an alloy
6. To provide data on the properties of the alloys after they have received one or more of the thermal and mechanical treatments described above
7. To determine the effect of modifications in heat-treatment practices that are suggested or investigated to overcome various production or service problems
8. To correlate mechanical properties with various other factors such as phase changes, precipitation of intermetallic compounds, and similar phenomena in basic studies on the physical metallurgy of the alloys.

A very large quantity of data has been accumulated to satisfy these objectives, and it is outside the scope of this report to attempt a comprehensive discussion for each alloy. Instead, tabulations and figures illustrate the effect of standard thermal and mechanical treatments on several mechanical properties. Also, the effects of certain modifications in treatment have been included for some of the more complex alloys to indicate the changes in properties that may be expected under certain circumstances.

In the following sections, the selected information is presented separately for each alloy. The extent and quantity of data presented depends on the form of the alloy (whether wrought or cast), on the number of thermal and mechanical operations that are normally used for the alloy, and on the degree of interest and usefulness of the alloy in industry.

NICKEL

The data presented in tables 4 to 11 and in figures 4 and 5 illustrate the differences in the mechanical properties of nickel resulting from various degrees of cold work, compared with the annealed and hot-worked material. That the tensile properties of nickel can vary over a wide range depending on its cold- or hot-worked condition is shown in table 4. As might be expected, tensile strength, yield strength, and hardness are increased by increasing the degree of cold work, while the ductility is decreased. Some ductility is retained, however, in the cold-worked and spring tempers. The hardness ranges for various mill products in relation to their condition are shown in table 8. In table 9, it can be seen that nickel retains its tensile properties up to about 600° F. The good properties at cryogenic temperatures are indicated in tables 10 and 11. These data show that the toughness of nickel in the annealed, hot-rolled, or cold-drawn conditions, as measured by the Charpy-V notch test, remains essentially unchanged down to a temperature of −300° F, with little effect attributed to the condition of the material. Similar results from other tests at very low temperatures are shown in figures 4 and 5 (ref. 24).

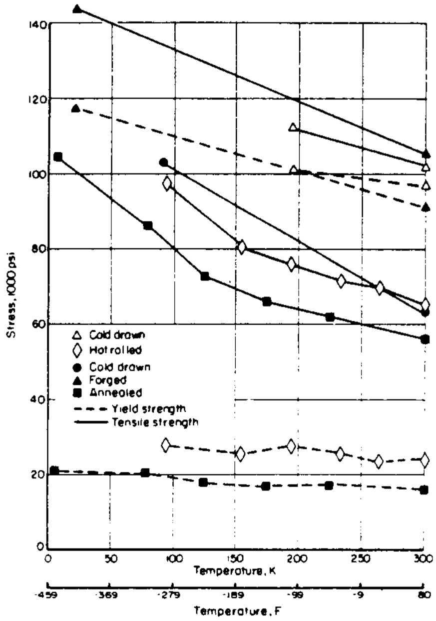

FIGURE 4.–Tensile and yield strength of nickel in various conditions at cryogenic temperatures.

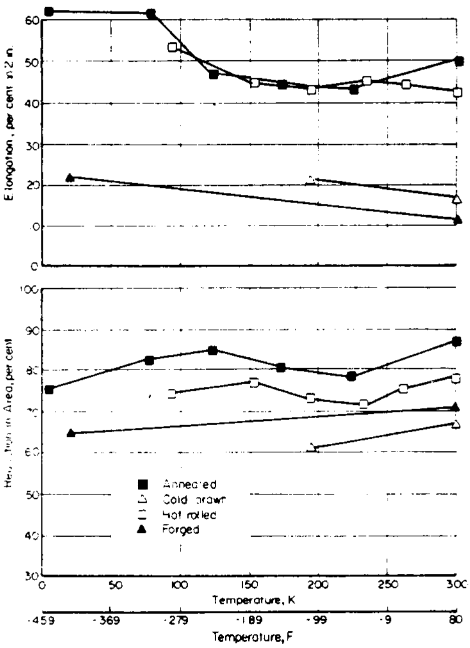

FIGURE 5.–Elongation and reduction in area of nickel in various conditions at cryogenic temperatures.

MONEL K-500

Monel K-500 is an age-hardenable nickel-copper alloy containing aluminum and titanium. Particles of $Ni_3(Al,Ti)$ precipitate throughout the matrix during the age-hardening heat treatment to develop increased strength and hardness. The alloy is used in applications where considerable strength is required in addition to corrosion resistance (ref. 2).

Annealing

Annealing to soften the matrix after cold working may be achieved with temperatures as low as 1400° to 1600° F, but heating at 1800° to 1900° F (a solution treatment) is recommended for optimum response to subsequent age hardening (ref. 25). For work to be age-hardened, it is important to water quench quickly from the heating temperature to avoid partial precipitation of the age-hardening phase. The effect on hardness of water quenching from temperatures below the optimum solution-treatment temperature is shown in figure 6. The increased hardness obtained if the alloy is heated at the lower temperatures, or if allowed to cool to such temperatures before quenching, reflects the effect of precipitation of the age-hardening constituents in the matrix. The approximate time at various temperatures required to soften the material is shown in figure 7. In this and other heating operations for Monel K-500, the furnace time should be the minimum required to heat the piece to temperature throughout. Prolonged heating has a detrimental effect on the ductility (ref. 17).

Age Hardening

The following procedures are recommended for obtaining maximum properties (ref. 25).

Soft material (140-180 Brinell, 75-90 Rockwell B).–Hold for 16 hr at 1080° to 1100° F followed by furnace cooling at the rate of 15° to 25° F/hr to 900° F. Cooling from 900° F to room temperature may be carried out by furnace or air cooling, or by quenching, without regard for cooling rate. This procedure is suitable for as-forged and quenched or annealed forgings, for annealed or hot-rolled rods and large cold-drawn rods (over 1-1/2-in. diam), and for soft temper wire and strip.

Moderately cold-worked material (175-250 Brinell, 8-25 Rockwell C).–Hold for 8 hr or longer at 1080°

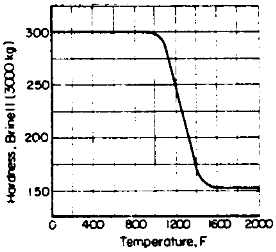

FIGURE 6.–Effect of water quenching from various temperatures on hardness of Monel K-500 (ref. 25).

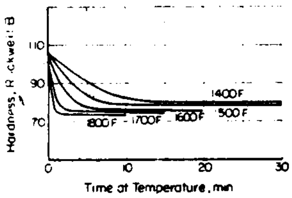

FIGURE 7.–Approximate time required at various temperatures to produce different degrees of softness in Monel K-500 by open annealing (ref. 18). (Quenching must follow annealing.)

to 1100° F, then cool to 900° F at a rate not to exceed 15° to 25° F/hr. Higher hardnesses can be obtained by holding for as long as 16 hr at temperature, particularly if the material has been cold-worked only slightly. As a general rule, material with an initial hardness of 175-200 Brinell should be held the full 16 hr. Material close to the top figure of 250 Brinell (25 Rockwell C) should attain full hardness in 8 hours. These procedures are applicable to cold-drawn rods, half-hard strip, cold-upset pieces, and intermediate-temper wire.

Fully cold-worked material (260-325 Brinell, 25-35 Rockwell C).–Hold for 6 hr or longer at 980° to 1000° F then cool to 900° F at a rate not exceeding 15° to 25° F/hr. In some instances slightly higher hardnesses may be obtained (particularly with material near the lower end of the hardness range) by holding 8 to 10 hr at temperature. This procedure is suitable for spring temper strip, spring wire, or heavily cold-worked pieces such as small, cold-formed balls.

Room-Temperature Properties

The nominal room-temperature mechanical properties for various forms of the alloy heat-treated according to these procedures are summarized in table 12, which also includes data illustrating the effect of cold or hot working, and of annealing. Cold working or aging increase the strength appreciably in all forms of the alloy. Still higher strength may be achieved by cold working prior to age hardening. The effect of cold work and of cold work plus age hardening on the tensile strength is shown in figure 8 (ref. 28).

Modifications of the aging procedures are possible to shorten the time or to obtain intermediate mechanical properties. Information that may be used as a guide for establishing suitable procedures is given in figure 9 and table 13.

Room-temperature data on other properties of Monel K-500–shear strength, bearing strength, compressive strength, torsional properties, impact strength, and fatigue resistance–are given in tables 14 through 19. Each gives the respective properties of the alloy in several conditions, illustrating the influence of various thermal and mechanical treatments.

High-Temperature Properties

Short-time high-temperature properties of rod are summarized in table 20. The data show the effect of age hardening hot-rolled rod. Tensile and yield strengths of the age-hardened material are considerably higher than those of the as-rolled material at

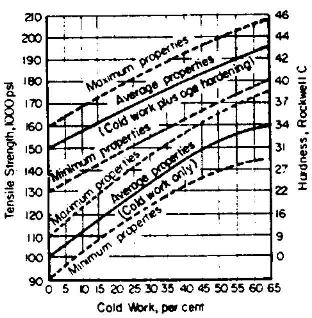

FIGURE 8.–Effect of cold work and age hardening on strength of Monel K-500.

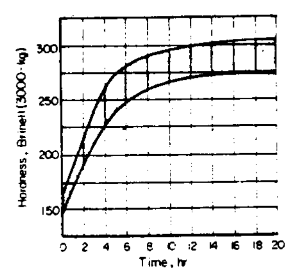

FIGURE 9.–Effect of time at temperature (1080° to 1100° F) on hardness of Monel K-500 (ref. 25). (Shaded area indicates range for commercial hot-rolled rods and forgings.)

temperatures up to about 1000° F. Likewise, up to at least 600° F, the strength of hot-rolled and aged material is slightly higher than that of the alloy when cold-drawn, annealed, and then age-hardened. Stress-rupture and creep-strength data are given in tables 21 and 22, and elevated-temperature fatigue-strength and hot-hardness data in tables 23 and 24.

Low-Temperature Properties

Monel K-500 has good low-temperature properties, allowing it to be used at temperatures as low as that of liquid hydrogen. The tensile properties of the alloy in the cold-drawn, and cold-drawn and aged conditions at temperatures down to −423° F, are shown in figures 10 and 11. Tensile and yield strengths for the cold-drawn material increase with a decrease in temperature, while the elongation remains essentially constant. The same is true for the age-hardened material although in this case results are shown only for temperatures down to −100° F. Additional data for age-hardened 0.020-in. sheet are given in table 25, showing results obtained to −423° F. The elongation and notched-unnotched ratios are higher at −423° F than at room temperature. Results of tests conducted at the Marshall Space Flight Center (ref. 27) for this alloy are shown in figure 12. Annealed and age-hardened 0.063-in. sheet specimens, both longitudinal

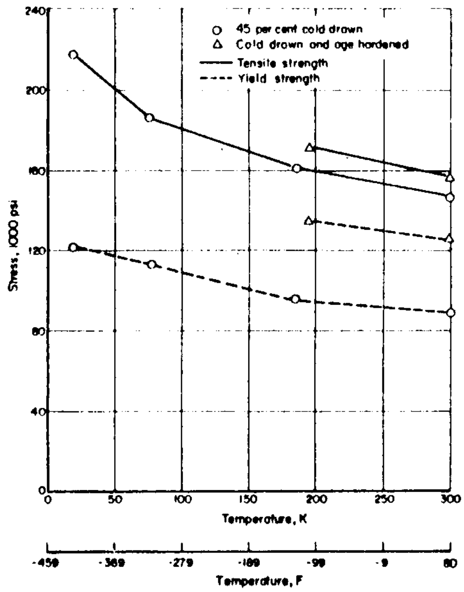

FIGURE 10.—Tensile and yield strengths of Monel K-500 at cryogenic temperatures (ref. 24).

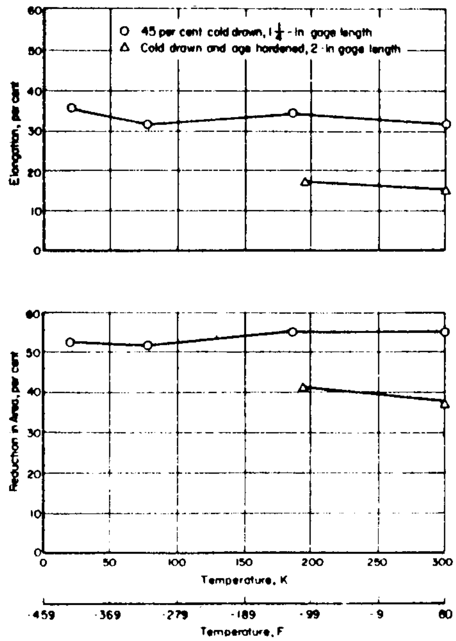

FIGURE 11.—Elongation and reduction in area of Monel K-500 at cryogenic temperatures (ref. 24).

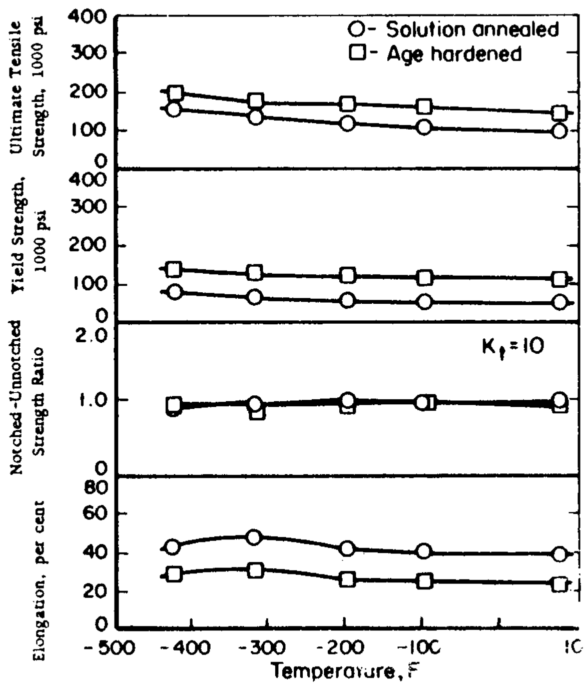

FIGURE 12.—Tensile properties of Monel K-500 at cryogenic temperatures (ref. 27). (0.063-in. sheet annealed and aged 16 hr at 110° F.)

and transverse to the rolling direction, were tested at temperatures from room temperature to −423° F. Figure 12 shows that the tensile and yield strengths of age-hardened material are higher than those of the annealed specimens at all temperatures, while the elongation is somewhat lower. However, in both conditions, tensile and yield strengths increased with decreasing temperature, while the elongation remained virtually unaffected.

Tension and bending impact strengths have been determined on smooth and notched specimens down to −200° F. The results are tabulated in table 26 and 27 (ref. 25). All test pieces had ductile fractures.

Decrease in temperature also increases fatigue strengths as shown in table 28. The material used was 0.051-in. sheet—cold-rolled half-hard, and aged, with a tensile strength of 182,000 psi. Tests were in flexure ($R = -1$) at 1800 cpm except those at −423° F, which were 3450 cpm. (See ref. 25.)

INCONEL X-750

Inconel X-750 is an age-hardenable nickel-chromium-iron alloy containing the aluminum and titanium required to form the hardening component, $Ni_3(Al, Ti)$. The nature and composition of the chromium carbides that are formed during heat treatment influence the mechanical properties. The alloy is useful in applications requiring corrosion and oxidation resistance, and has high creep-rupture strength at temperatures up to about 1500° F. It retains its properties to liquid-hydrogen temperature and may be used for cryogenic applications. Various heat treatments are employed, depending on the application and properties desired.

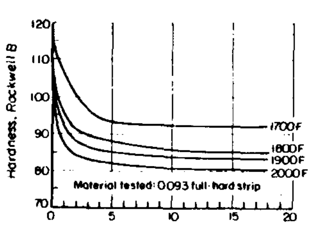

FIGURE 13.—Approximate time required at various temperatures to produce different degrees of softness in Inconel X-750 by open annealing (ref. 18). (Quenching must also follow annealing.)

Annealing

For maximum softening, bar stock is generally mill annealed at 2000° F for 45 min and air-cooled. The hardness of material after heating at this and other temperatures is shown in figure 13. Relatively high strength is retained after annealing at lower temperatures. Typical room- and elevated-temperature tensile properties of round bar annealed at 1750° F are given in table 29.

Age Hardening

Several standard heat treatments have been recommended for various forms of the alloy (refs. 28 and 29). Some of these are designed to develop high creep and rupture strengths at temperatures above 1100° F. Others are suggested for lower service temperatures. Modifications have been developed that reduce the heating time, while producing satisfactory strength properties. The conditions for several of these treatments are given in table 30. The effects of these treatments as well as some experimental procedures on the mechanical properties, and some comparisons between results obtained by the different treatments, are given in the tables and figures that follow.

Room-Temperature Properties

The room-temperature tensile properties of hot-rolled and stress-equalized bar, aged at 1300° F, are given in table 31. Similar results for cold-rolled and annealed sheet aged at various temperatures and for several time periods are given in table 32. Some of these data are plotted in figure 14. The 1300° F/20-hr treatment (treatment 8, table 30) develops the highest strength properties upon direct aging of mill-annealed material. It has been reported (ref. 28), however, that the 1400° F/1 hr, FC treatment (treatment 7, table 30) will develop approximately equal strength. Another comparison of heat treatments on the room-temperature tensile properties of hot-rolled bars is seen in table 33. This illustrates that a shorter aging cycle (total heating time, 7 hr vs 17 hr) may be employed if slightly lower strength can be tolerated. A similar comparison for cold-rolled, annealed sheet over a thickness range from 0.012 to 0.140 in. is shown in table 34. Higher strengths are obtained by furnace cooling than by air cooling from the aging temperature.

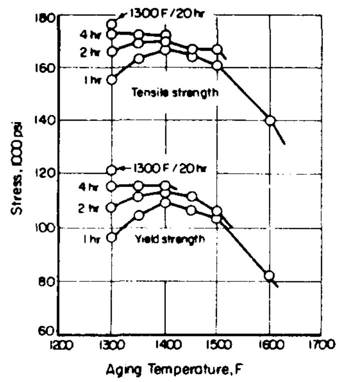

FIGURE 14.–Effect of aging conditions on room-temperature tensile properties of annealed Inconel X-750 Sheets (ref. 28).

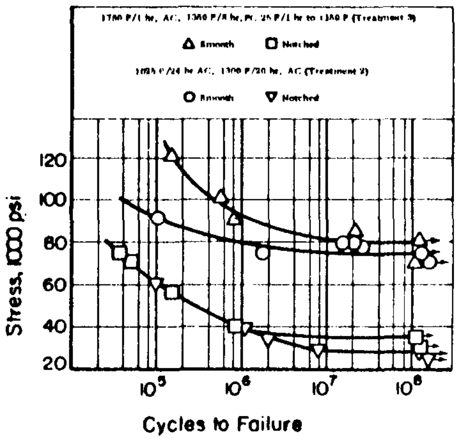

FIGURE 15.–Effect of heat treatment on fatigue life of 3/4-in. Inconel X-750 hot-rolled bar.

The effect of heat treatment on fatigue life at room temperature is shown in figure 15 (ref. 28). Samples were tested in completely reversed bending.

Cold working also increases the tensile strength, and subsequent aging increases strength still more. This is shown in figure 16 (ref. 18), and the results of straining sheet material on the room-temperature tensile properties are tabulated in table 35. The effects of several different heat treatments on the properties of wire in two tempers are summarized in table 36.

Similarly, hot working of the alloy as in forging also results in increased room-temperature yield and

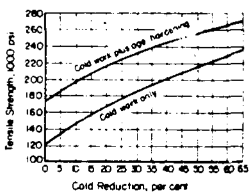

FIGURE 16.–Effect of cold work and age hardening on Inconel X-750.

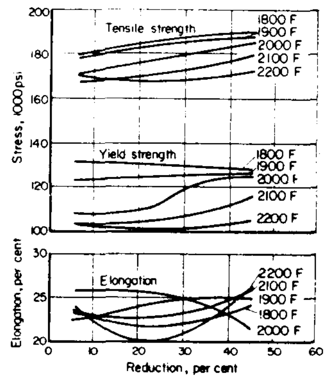

FIGURE 17.–Effect of forging temperature and amount of reduction on room-temperature tensile properties of Inconel X-750 (ref. 23). (Specimens aged by treatment 2 in table 30.)

tensile strengths. This is shown in figure 17, which also indicates the influence of forging temperature within the normal 1800° to 2200° F range.

High-Temperature Properties

Considerable data have been accumulated on the high-temperature properties of Inconel X-750 sheet and bar. The data presented in the following tables show comparative effects of various thermal and mechanical treatments.

Table 37 summarizes the effect of two age-hardening procedures on the hot hardness of bar. Direct aging of the bar (treatment 8, table 30) results in a higher hardness at all temperatures than that given by a solution and double-aging treatment (treatment 1, table 30).

The high-temperature tensile properties of sheet and bar stock age-hardened by several procedures are given in tables 38 and 39 and in figure 18. The effect on stress-rupture life of bar stock is illustrated in figures 19 and 20.

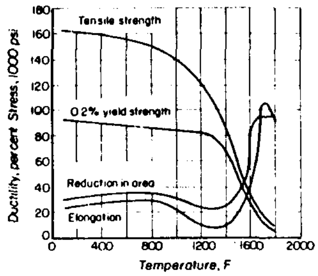

FIGURE 18.–Room- and elevated-temperature tensile properties of Inconel X-750 bar stock (ref. 28). [Heat treatment: 2100° F/2 hr, AC; 1500° F/24 hr, AC; 1300° F/20 hr, AC (treatment 1, table 30.)]

Furnace cooling from the aging temperature results in better strength of sheet at high temperature than is obtained by air cooling. This was also true for room-temperature tensile properties (see table 34).

The increase in yield and tensile strengths for hot-rolled bars (with only a slight decrease in tensile ductility in the 600° to 1000° F range and at 1500° F) shown in table 39 is attributed to a combination of higher solution-annealing temperature and slow furnace cooling through the temperature range of 1350° to 1150° F. Solution annealing at 1750° F for 1 hr rather than at 1625° F for 24 hr (stress equalizing) increases strength after aging because of: (1) less softening due to annealing of previously worked structure, and (2) more gamma prime in solution, so that on subsequent aging more hardening phase or gamma prime, $Ni_3(Al, Ti)$, is precipitated.

The tensile strength properties for the 1350° to 1150° F aging treatment are higher than those developed by the triple heat treatment (age at 1550° and at 1300° F) or by equalizing and aging at 1300° F (compare values in table 39 with those shown in figure 18). A comparison of figures 19 and 20 shows the superior rupture life attained with the triple heat treatment, in comparison with equalized and aged material, above approximately 1100° F.

The effect of prolonged exposure at elevated temperature under stress (1000 hr, 40,000 psi, 650° F) on the tensile properties of 0.025-in. cold-rolled sheet was determined in a study of materials for supersonic transports (ref. 31). The results, summarized in table 40, indicate relatively little effect on properties.

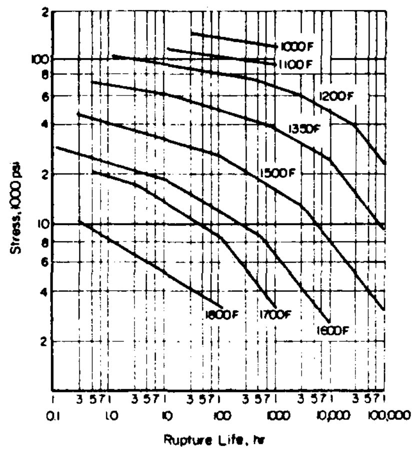

FIGURE 19.–Rupture strength of Inconel X-750 bar stock (ref. 28). [Heat treatment: 2100° F/2 hr, AC; 1550° F/24 hr, AC; 1300° F/20 hr, AC (treatment 1, table 30.)]

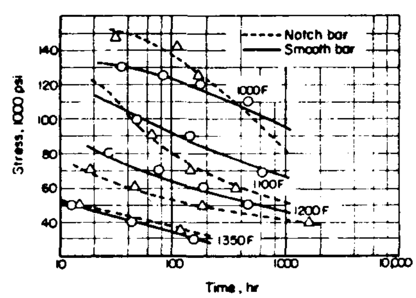

FIGURE 20.–Effect of notch on rupture life of Inconel X-750 bar stock (ref. 28). [Heat treatment: 1625° F/24 hr, AC; 1300° F/20 hr, AC, smooth bar, 0.3-in. diam x 1½-in. long; notched bar, 50%, 60-deg V notch, 0.005-in. root radius (treatment 2, table 30.)]

Sheet that had been cold-rolled 67% was appreciably stronger than that cold-rolled only 27 percent.

The influence of heat treatment on impact strength and on high-temperature fatigue strength is shown in table 41 and figure 21 (ref. 28), respectively.

That the forging temperature and amount of reduction definitely affect the stress-rupture life of Inconel X-750 bars is shown in figure 22. (Heat treatment: 2100° F/2 hr, AC; 1550° F/24 hr, AC; 1300° F/20 hr, AC; test conditions: temperature, 1350° F, stress, 45,000 psi.)

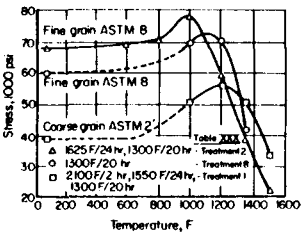

FIGURE 21.–Effect of heat treatment and resulting grain size on fatigue strength of Inconel X-750 (10^3 cycles). (Samples were tested in completely reversed bending.)

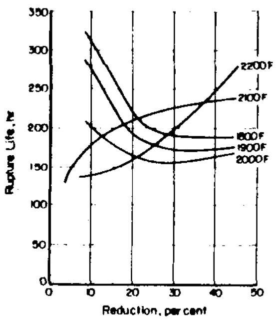

FIGURE 22.–Effect of forging temperature and amount of reduction on the rupture life of Inconel X-750 forged bars (ref. 28).

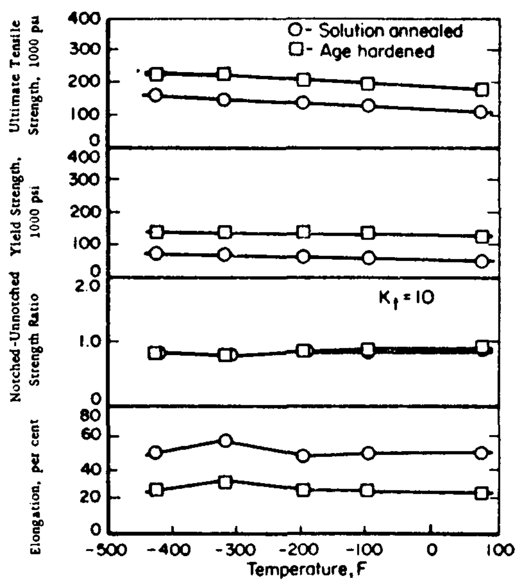

FIGURE 23.–Tensile properties of Inconel X-750 at cryogenic temperatures (ref. 27). (0.063-in. sheet, annealed and aged 20 hr at 1300° F.)

Low-Temperature Properties

The results of several investigations to determine the mechanical properties of sheet and bar at low temperatures are summarized in tables 42 and 43 and figure 23. They show that Inconel X-750 has good mechanical properties down to –423° F and can be considered for cryogenic applications. The only comparison showing the influence of heat treatment is that given in figure 23. It shows that the superior tensile and yield strengths of the age-hardened material are retained at low temperatures, while the elongation was reduced by about 1/2 of the annealed values, but still remained over 20% at all temperatures.

ALLOY 718

Alloy 718* is a precipitation-hardenable alloy having good mechanical properties at temperatures from –423° to 1300° F. It differs from most nickel-base superalloys in that it is strengthened by a gamma-prime precipitate containing a considerable amount of columbium (ref. 32). A comparison of the nominal chemical compositions of the selected alloys (table 1) shows that Alloy 718 contains much less aluminum and titanium than the other alloys but, instead, contains appreciably more columbium. In early studies of the alloy, Eiselstein (ref. 33) showed that maximum yield strength was developed on aging alloys containing about 50% to 55% nickel, with columbium in the range from 4.5 to 6.5% (fig. 24). The lower curves for the alloys in the annealed condition indicate that columbium had little solid-solution-strengthening effect. Figure 25 shows that the strength after aging is still higher with columbium contents up to 8%, but this was reported to cause considerable reduction in ductility.

Many attempts have been made to relate the microstructure and phase compositions to the properties of Alloy 718 (refs. 34 to 38). Early work showed that a gamma prime phase, $Ni_3(Cb, Al, Ti)$, was the primary hardening constituent. Raymond and Cometto (ref. 23) found that gamma prime is a disc shaped, ordered body-centered-tetragonal phase coherent with the face-centered-cubic matrix in the age-hardened condition. As indicated by figures 26, 27,

*Alloy 718 is manufactured and marketed under various trade names. The original name given to it by the developer, International Nickel Co., is Inconel 718, and many people still refer to it by that name.

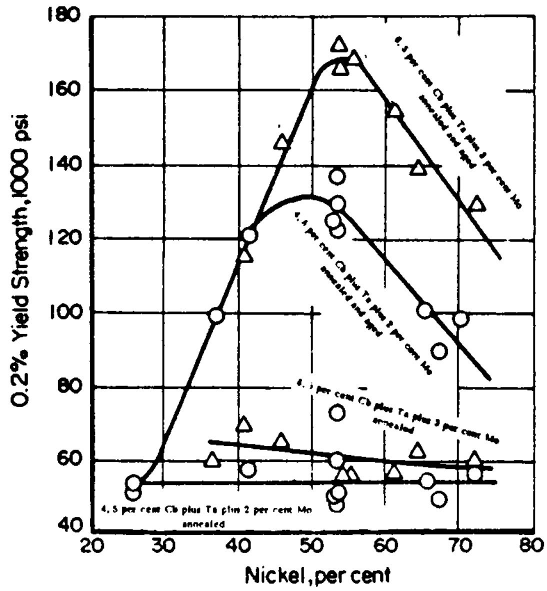

FIGURE 24.–Effect of nickel and columbium plus tantalum content on the room-temperature yield strength of Ni, Cr, Cb, Mo, Al, Ti, and Fe alloys (ref. 33). (Annealed at 1900° F/1 hr, WQ plus aged at 1250° to 1350° F/16 hr, AC, and annealed at 1900° F/1 hr, WQ, no aging treatment.)

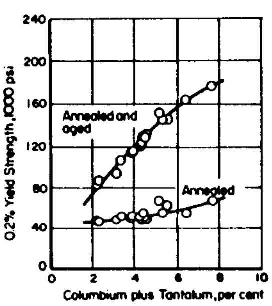

FIGURE 25.–Effect of percent columbium plus tantalum on 0.2 percent yield strength (ref. 33). (Material was annealed at 1900° F/1 hr, WQ, aged at 1250° to 1350° F/16 hr, AC.)

28, and elsewhere (refs. 38 and 39), the phases commonly found in commercially heat-treated Alloy 718 are MN, M(NC), MC, Ni_3Cb (orthorhombic), M_6C, Ni_2Cb (Laves), gamma prime, and more recently Ni_xCb, Ti_2 (S, C), and "J" phase. The relationship between these phases is very dependent on prior processing, heat treatment, composition, strain, and other variables.

Studies have been made to determine the optimum composition of Alloy 718, and some results will be given later. However, the optimum composition of the alloy is dependent on the properties desired for specific applications. Furthermore, chemical composition, heat treatment, and mechanical properties are

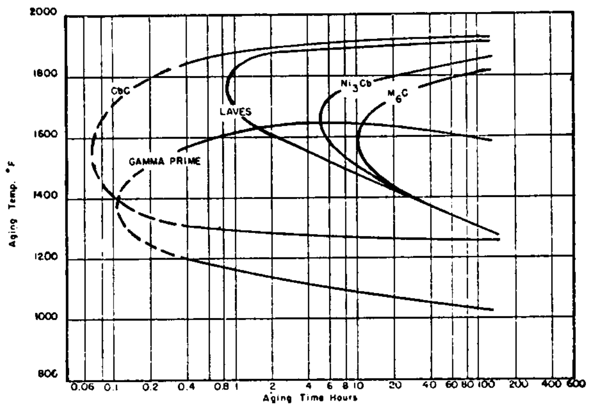

FIGURE 26.–Time-temperature-transformation diagram of air-melted Alloy 718 rod after annealing 2100° F/1 hr, WQ (ref. 39).

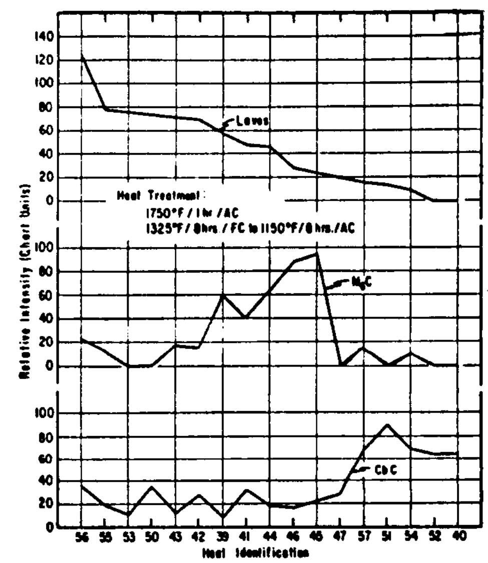

FIGURE 27.–Relative amounts and relationship of Laves, M_6C and CbC in experimental Alloy 718 heats (ref. 38).

interrelated, and a number of changes have been made since the alloy was first introduced. Wagner and Hall (ref. 13) reviewed the developments that have taken place and reported that specifications on chemical composition have become more restrictive than indicated by the nominal composition. The effects of the chemical composition on properties of the alloy are dependent on the heat-treatment conditions. Therefore, a number of variations in the annealing and age-hardening procedures have been recommended or

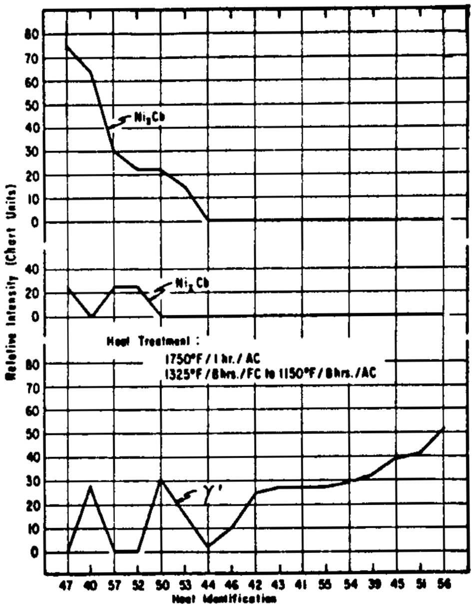

FIGURE 28.—Relative amounts and relationship of Ni_3Cb,Ni_xCb, and gamma prime in experimental Alloy 718 heats (ref. 38).

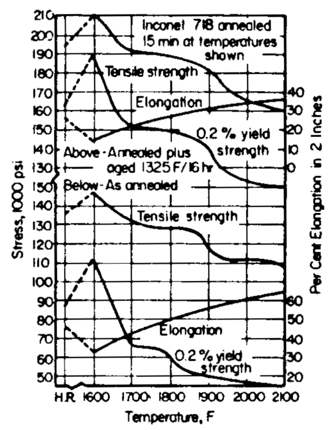

FIGURE 29.—Effect of annealing temperature on the room-temperature tensile properties of Alloy 718. (Lower curves show as-annealed condition; upper curves show effect of age hardening. H.R. = hot rolled.)

specified, taking these interrelationships into account. The properties that may be obtained with some of these combinations of chemical composition and heat treatment are illustrated in subsequent sections.

Effect on Room-Temperature Properties

Studies of annealing conditions for Alloy 718 have usually been concerned with the mechanical properties of the alloy after various aging treatments. Therefore, it appears desirable to discuss the annealing and age hardening operations together, rather than separately. Data showing the effects of cold working are often shown in the same figures and tables with the effects of heat treatment. Therefore, cold-working information is also included in this section.

In the original recommendations for this alloy (ref. 40), annealing at about 1750° F, followed by a single aging step of 16 hr at 1325° F, was recommended for mill-annealed products. The effect of various annealing temperatures on room-temperature tensile properties, as annealed and after age hardening, is shown in figure 29 (ref. 40). Experiments reported by Barker (ref. 32) indicated that changes in aging conditions could result in appreciably higher yield strengths. The possibilities in this direction are shown by figure 30. The 200-hr, 1200° F second aging step is not considered practical, but the results were obtained in connection with studies of simulated service conditions, and are shown for comparison. The "furnace cool age" curve compared with the "normal age" curve shows the improvement resulting from the use of a more practical two-step aging treatment. Other two-step aging treatments have been developed and typical ones are listed in table 44. These differ not only in aging conditions, but also in the annealing temperature. Increasing the annealing temperature up to 1950° F originally was recommended to obtain better creep and rupture strengths. As will be shown later, however, the lower notch strength at 1200° F is a disadvantage of annealing at a higher temperature. The data in table 45 show the effect of increased annealing temperature on the room-temperature tensile properties. Data from another source are plotted in figure 31. (The data were obtained with specimens machined from upset forgings. The following aging treatment was used: 8 hr at 1325° F, cool to 1150° F at 100° F/hr, hold 8 hr at 1150° F, AC.) Eiselstein (ref. 33) pointed out that for applications which are tensile rather than creep limited, annealing at 1950° F

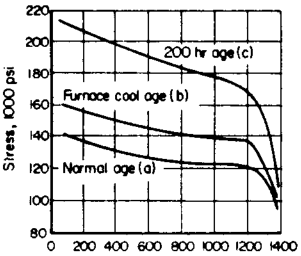

FIGURE 30.–Comparison of yield strengths of Alloy 718 sheet obtained by different aging treatments (ref. 32). (Solution treated 1700° F, 1 hr, AC, then: (a) 1325° F, 16 hr, AC; (b) 1325° F, 8 hr, FC 20° F/hr to 1150° F, AC; and (c) (a) plus 1200° F, 200 hr.)

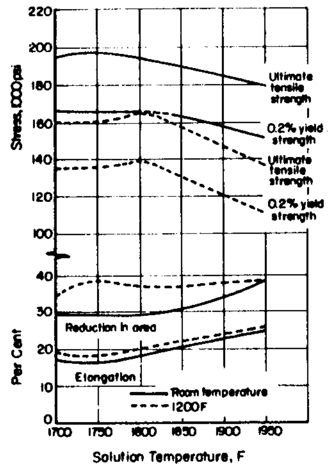

FIGURE 31.–Effect of solution temperature on the tensile properties of subsequently aged Alloy 718 (ref. 41).

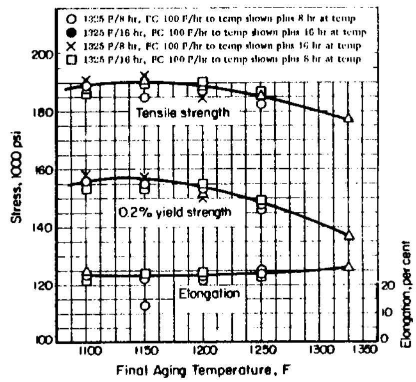

FIGURE 32.–Effect of final aging temperature on room-temperature tensile properties of 0.060-in. cold-rolled Inconel 718 sheet annealed 1700° F/1 hr (ref. 43).

is nevertheless required to fully soften the material for severe cold-forming operations. Aging temperatures must then be increased (as shown in table 44) to permit optimum age hardening. Newcomer (ref. 16) investigated the annealing temperature necessary to redissolve the precipitate in overaged material so that subsequent aging would produce maximum hardness. Microstructure and grain size were also observed. The data showing the effect of annealing temperature on hardness and grain size are given in table 46. According to the author, these data along with metallographic examination indicated that annealing at 1500° or 1600° F failed to dissolve the precipitated phases and did not restore the hardness to that of the mill-annealed material. Annealing between 1700° and 1800° F adequately dissolved the precipitate so that subsequent aging produced maximum hardness, but annealing at 1900° F was necessary to completely dissolve all the precipitates within the grains and at the grain boundaries. Heating above 1900° F caused excessive grain growth.

The curves in figure 32 indicate how tensile properties can vary with the temperature of the second aging. The best combination of properties seems to develop with a final aging temperature of 1150° F on sheet annealed at 1700° F.

The interrelationship between chemical composition, heat treatment, and mechanical properties was mentioned earlier. Data and curves from Eiselstein reproduced by Wagner and Hall (ref. 13) in figures

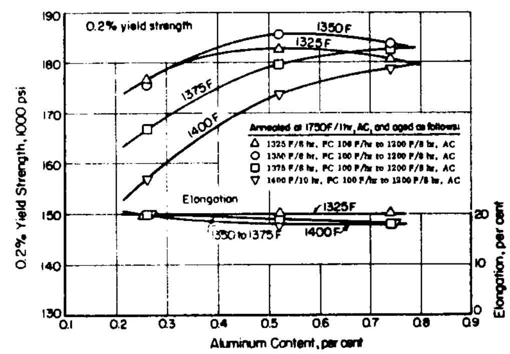

FIGURE 33.–Effect of aluminum content on the room-temperature yield strength of Inconel 718 hot-rolled bar stock (ref. 13).

33 and 34, show the change in room-temperature tensile properties as influenced by aluminum content and heat treatment. Highest yield strengths are obtained with a low-aluminum content, a 1950° F anneal, and aging at 1325° F. These conditions, however, are said to cause notch brittleness in creep testing.

The increase in tensile and yield strengths that may be achieved by cold working prior to aging is shown in figures 35 to 37. The elongation is decreased proportionately, but is still about 5% after the maximum cold reduction.

One of the unique characteristics of Alloy 718 is its sluggish response to aging. This permits annealing

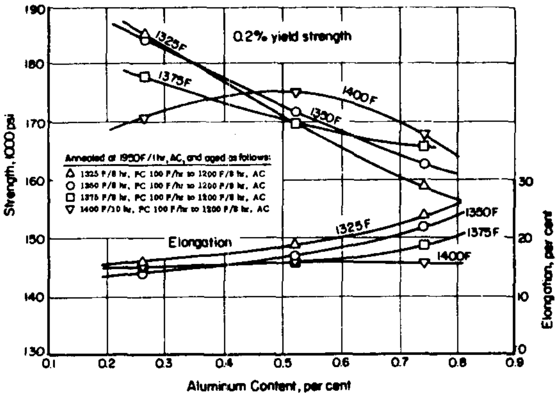

FIGURE 34.–Effect of aluminum content on the room-temperature 0.2% yield strength of Inconel 718 hot-rolled bar stock (ref. 13).

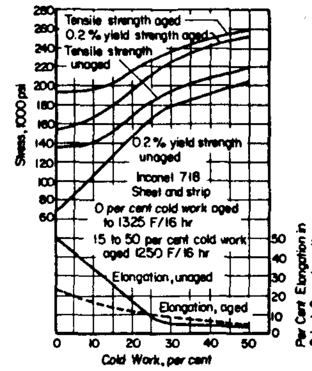

FIGURE 35.–Tensile properties of Alloy 718 vs percent cold work (ref. 40).

and welding without spontaneous hardening during heating and cooling. The effect of aging time and temperature on the hardness of mill-annealed sheet is shown in figure 38 (ref. 40).

Rizzo and Buzzanell (ref. 38) studied the effect of composition and heat treatment on the tensile and impact properties of Alloy 718 that had been double vacuum melted and hot-rolled to 1/2-in. bars. Annealing at 1750° F prior to aging gives higher yield and ultimate strength than annealing at 1950° F, with a concomitant lower ductility (figs. 39 and 40). Figure 41 shows that the impact strength correlates with the tensile ductility; both are greater with the higher annealing temperature at 1950° F. The yield strength appears more sensitive to variations in composition, especially columbium content, than does the ultimate strength.

Multiple regression analyses of the tensile and impact properties, in which the four variations in composition (table 47) and the two annealing temperatures were used, showed that columbium had the greatest effect on the properties. Columbium, silicon, and titanium improved tensile strength but decreased tensile and impact ductility. The effect of aluminum was too variable to categorize. Selected regression curves are shown in figures 42 through 45.

That increased amounts of Laves phase, Ni_3Cb, and gamma prime increased strength and reduced ductility is also noteworthy. Increased M_6C appeared detrimental to strength in material annealed at 1750° F but not when annealed at 1950° F.

In research at Benét Laboratories (U. S. Army Weapons Command), wrought billets of Alloy 718 were hydrostatically extruded at room temperature and heat-treated in an effort to improve the room-temperature properties (ref. 44). The strength of aged material increased while its ductility decreased with greater deformation as shown in figure 46. Optimum strength (300 ksi) of annealed and extruded (50% reduction in area) Alloy 718 is seen in figure 47 to result from the 1150° F/8-hr aging treatment. Tensile data for other treatments are given in table 48.

The effect of annealing temperature on double-vacuum-melted and vacuum-cast Alloy 718 is shown by the bar graph in figure 48 (ref. 45). Annealing of the cast and machined bars resulted in much lower strength, as expected, but subsequent aging caused an increase in strength with a corresponding drop in ductility. A significant improvement in both strength and ductility of the cast material was also obtained by annealing at 2150° F/2 hr, AC, compared to either oil quenching (OQ) or annealing at 1850° F prior to aging. These results are unlike those of wrought material, which is weaker with a higher annealing temperature prior to aging. Test bars that were cast-to-size exhibited lower strengths than test specimens that had been machined from 1/2-in. x 1-1/4-in. x 5-1/4-in. castings.

Compression data for the cast Alloy 718 is given in table 49 (ref. 45).

The effect of aging on the hardness of cold-drawn (64% reduction of area) wire is shown in table 50 (ref. 46). The maximum hardness is seen to occur at a lower aging temperature in the drawn wire than it does in annealed material.

Tilley (ref. 47) studied the effect of heat treatment on the tensile properties of Alloy 718 at room tem-

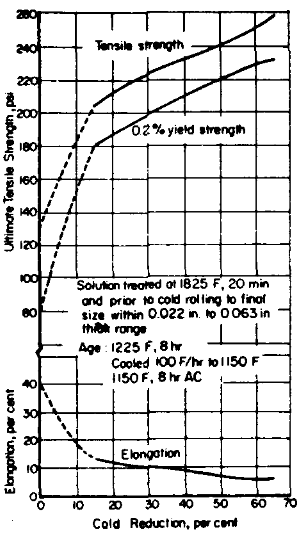

FIGURE 36.–Effect of cold reduction on mechanical properties of Alloy 718 sheet (ref. 41).

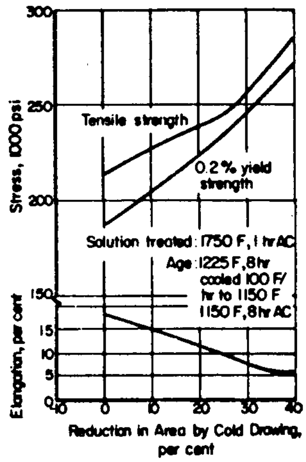

FIGURE 37.–Effect of cold reduction on mechanical properties of Alloy 718 bar stock (ref. 41).

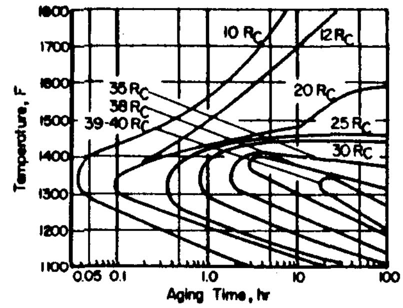

FIGURE 38–Effect of aging temperature and time on Rockwell C hardness of mill-annealed Alloy 718 sheet. (Initial hardness as annealed was RC4.)

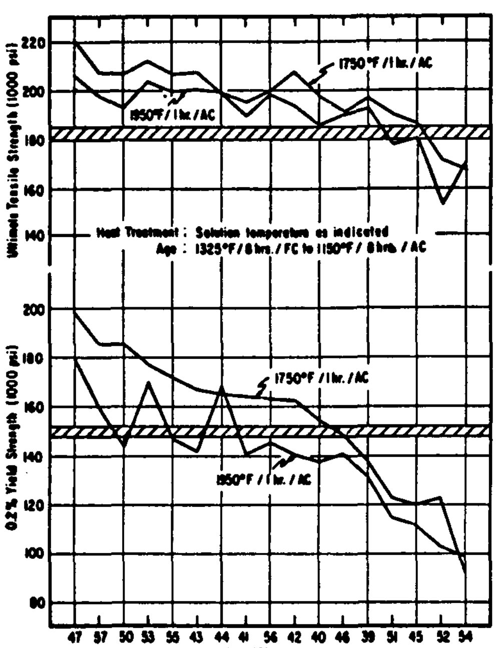

FIGURE 39.–Effect of solution treating temperature on room-temperature ultimate tensile and 0.2 percent yield strength of experimental Alloy 718 heats (ref. 38).

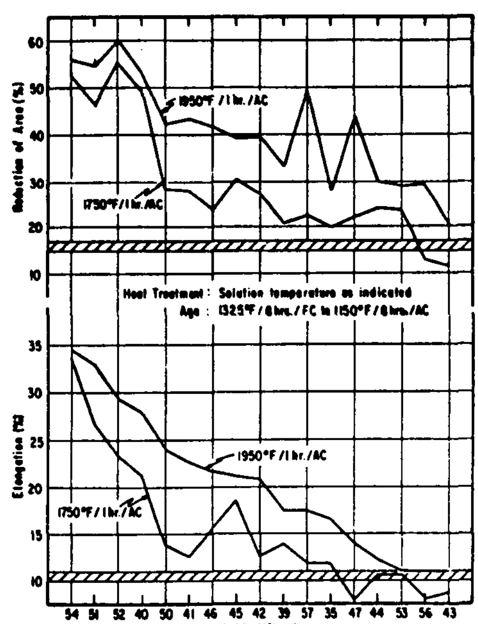

FIGURE 40.–Effect of solution treating temperature on the tensile ductility of experimental Alloy 718 heats (ref. 38).

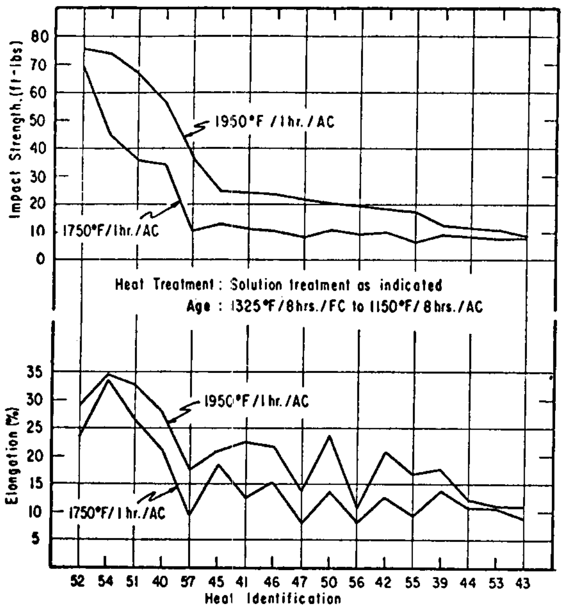

FIGURE 41.–Effect of solution temperature on the impact strength of experimental Alloy 718 heats and its relationship to tensile elongation (ref. 38).

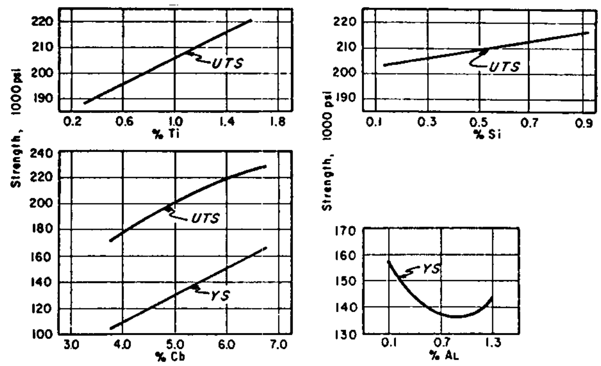

FIGURE 42.–Regression curves showing the effect of Cb, Ti, Si, and Al on the ultimate tensile and 0.2 percent yield strength of experimental Alloy 718 heats solution treated at 1950° F (ref. 38).

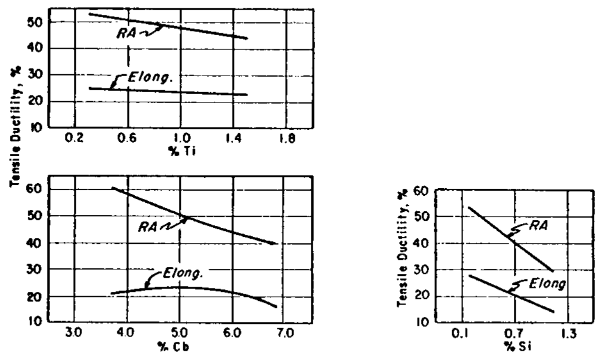

FIGURE 43.–Regression curves showing the effect of Cb, Si, and Ti on the tensile ductilities of experimental Alloy 718 heats solution treated at 1950° F (ref. 38).

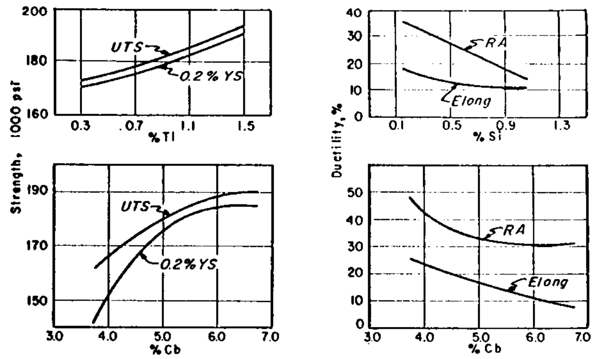

FIGURE 44.–Regression curves showing the effect of Cb, Ti, and Si on the tensile properties of experimental Alloy 718 heats solution treated at 1750° F and exposed for 1500 hr at 1200° F (ref. 38).

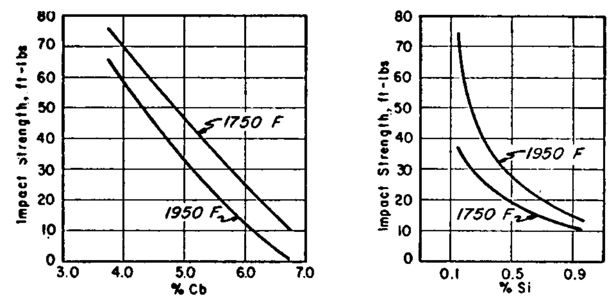

FIGURE 45.–Regression curves showing the effect of Cb, Si, and solution temperature on the impact properties of experimental Alloy 718 heats (ref. 38).

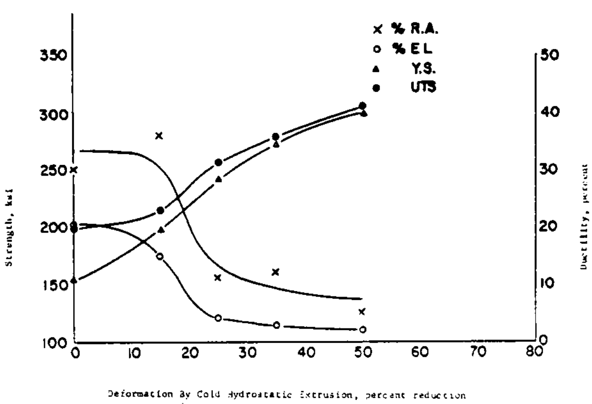

FIGURE 46.–Effect of hydrostatic extrusion on the room-temperature strength and ductility of Alloy 718 (ref. 44). (Billets were given 1800° F/4 hr, AC + 1325° F/8 hr, AC, cold hydrostatically extruded, then aged 1150° F/8 hr, AC.)

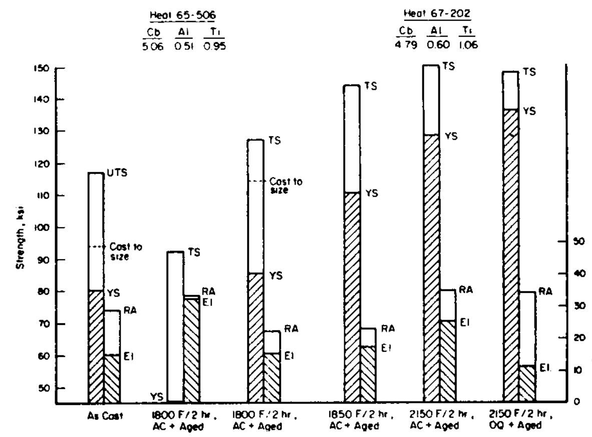

FIGURE 48.–Effect of annealing treatment on the room-temperature strength and ductility of as-cast (vacuum-melted and vacuum-cast) Alloy 718 when aged 1325° F/8 hr, FC to 1150° F/8 hr, AC.

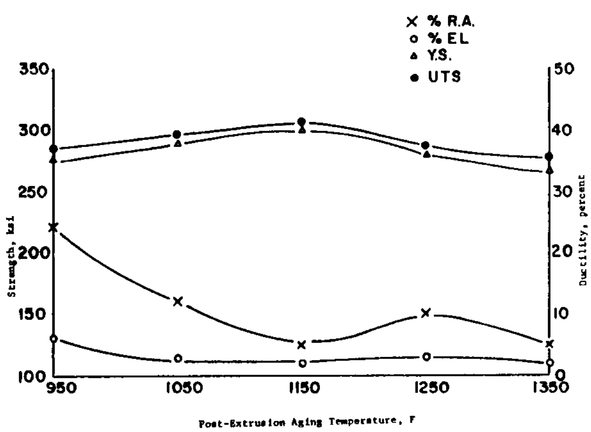

FIGURE 47.–Effect of aging temperature on the room-temperature tensile properties of wrought Alloy 718 (ref. 44). (Material was annealed at 1800° F/4 hr, AC, cold hydrostatically extruded 50 percent, aged at temperature shown for 8 hr.)

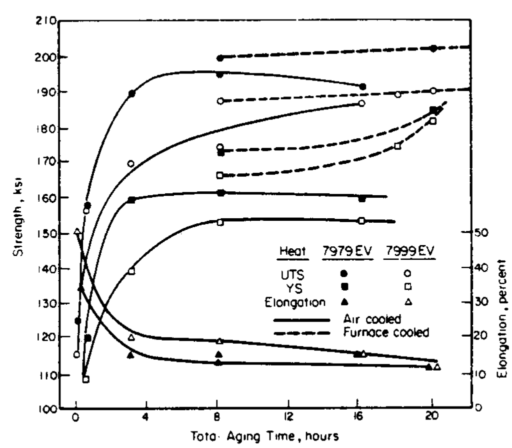

FIGURE 49.–Room-temperature tensile properties of vacuum-melted, wrought Alloy 718 annealed at 1750° F/1 hr, AC with various aging treatments.

perature, using small billets (1.875 in. x 5 in. and 6 in. x 12 in.) from two vacuum-arc-melted heats. Figures 49 through 52 (ref. 47) show the heat containing more columbium (7979) to be stronger and less ductile than the heat with a higher aluminum plus titanium content (7999). This tends to confirm that columbium is the element primarily responsible for the age hardening of Alloy 718. At each annealing treatment the strength increased with aging time while the ductility was reduced accordingly. The "two-step" aging treatments (aged at T_1, furnace cooled to T_2, held) resulted in stronger material which did not overage or soften as fast as in single-temperature aging. The aging treatments employed are given in table 51 (ref. 47). Faster overaging occurred after annealing at 1800° F. Even then, with the material being overaged at 1325° F/16 hr, AC, the longer two-step treatments resulted in significant yield-strength improvements. Annealing at 1950° F (and 1950° + 1750° F) permitted large jumps in yield strength when two-step

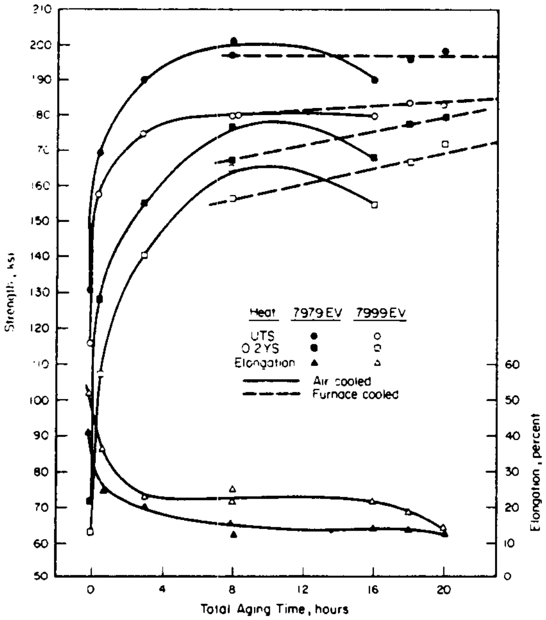

FIGURE 50.—Room-temperature tensile properties of vacuum-melted, wrought Alloy 718 annealed at 1800° F/1 hr, AC with various aging treatments.

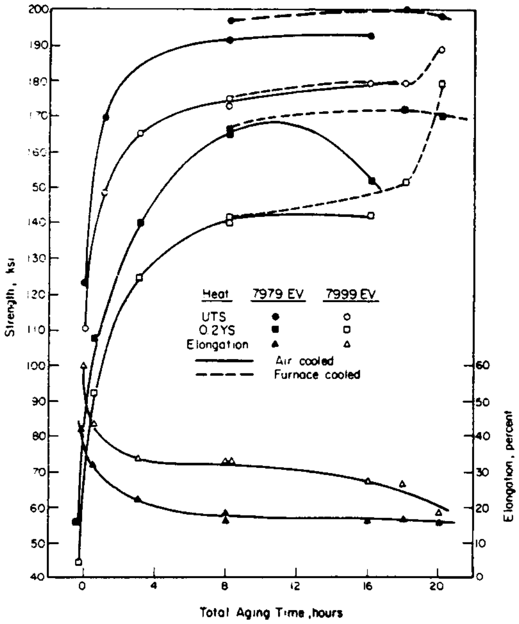

FIGURE 51.—Room-temperature tensile properties of vacuum-melted wrought Alloy 718 annealed at 1950° F/1 hr, AC with various aging treatments.

aged at 1400° and 1200° F (total of 20 hr). Aging at 1400° F/10 hr, FC to 1200° F for even longer than 10 hr may further increase the yield strength of Alloy 718 but might reduce the ductility to a prohibitively low level. The maximum ultimate strength of material annealed at 1750° F was only slightly higher than the maximum at the other annealing temperatures, but the maximum yield strength was about 10 ksi greater.

Additional data on the room-temperature properties of Alloy 718 may be found in references 39 and 48 to 51.

The impact strength of two heats of Alloy 718 are shown in figures 53 and 54 after various annealing and aging treatments. These data indicate improved toughness as a result of annealing at the higher temperature.

Effect on High Temperature Properties

The effect of chemical composition and the development of heat treatments were largely discussed in the preceding section. Considerable data have also been accumulated to show how mechanical properties at high temperatures are influenced by heat treatments. The tensile properties developed by the original one-step aging are shown in figure 55. The improvement in tensile and yield strengths, and the drop in ductility caused by cold working prior to aging, is shown by comparing figure 55 with figure 56. Additional data were obtained during the development of the two-step aging treatments and the improved properties attainable with various new procedures are shown in figures 57 and 58, and in table 52 (ref. 43).

Cullen and Freeman (ref. 52) determined the properties of Alloy 718 sheet at 800°, 1000°, and 1200° F to evaluate the potential usefulness of the material in the supersonic transport. Tensile properties were determined on smooth and notched ($K_t > 20$) specimens. Heat-treatment conditions were (1) cold-worked 20% and aged, (2) cold-worked, annealed 1 hr at 1750° F and aged, and (3) cold-worked, annealed 1 hr at 1950° F, and aged. The results of heat treatment are summarized in tables 53 to 55. In the cold-worked and aged condition, the notched/unnotched tensile strength ratio drops from about 1.0 at room temperature to 0.8 at 1200° F. Annealing at 1950° F followed by aging results in an essentially constant ratio of about 0.96 at all of the test temperatures.

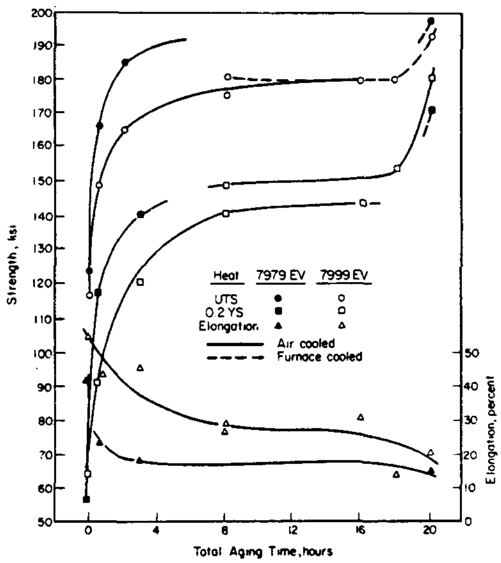

FIGURE 52.–Room-temperature tensile properties of vacuum-melted, wrought Alloy 718 annealed at 1950° F/1 hr, AC + 1750° F/1 hr, AC with various aging treatments.

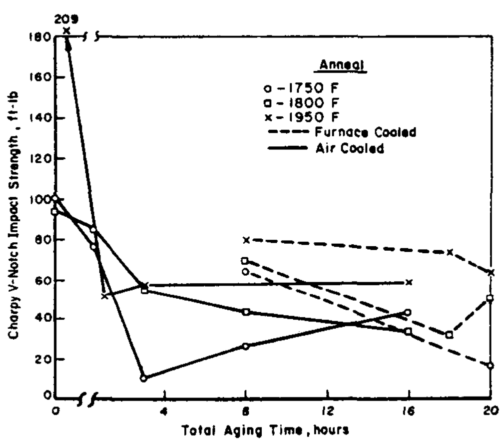

FIGURE 53.–Room-temperature impact strength of vacuum-melted, Alloy 718 billet showing the effect of annealing and aging treatments, heat 7999 (ref. 47).

Higher strengths are obtained by direct aging of cold-worked sheet. Additional data showing the same effects over a temperature range from – 320° to 1000° F are shown in table 56. This table also indicates that the properties are not affected by prior exposure for 1000 hr at a temperature of 650° F and a stress of 40,000 psi.

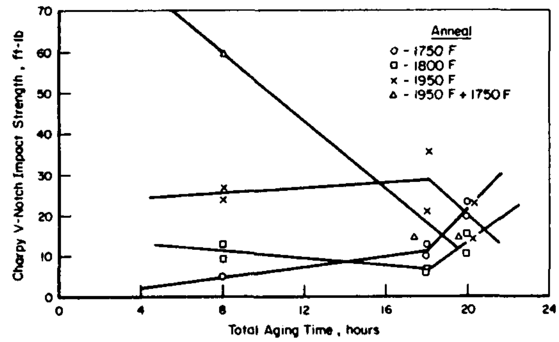

FIGURE 54.–Room-temperature impact strength of vacuum-melted, Alloy 718 billet showing the effect of annealing and aging treatments, heat 7979 (ref. 47).

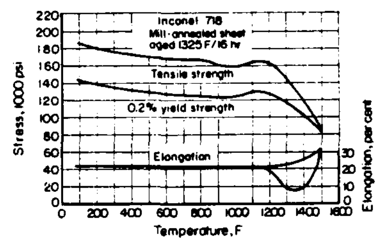

FIGURE 55.–Tensile properties vs temperature for Alloy 718 sheet, (ref. 40). (The elongation normally is as shown by the upper line in the 1200° to 1500° F range. Heating at temperatures above 1800° F, e.g., during annealing or furnace brazing, may reduce the elongation in this range to the values indicated by the lower line.)

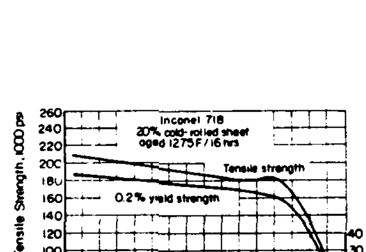

FIGURE 56.–Effect of cold work on elevated-temperature tensile properties of Alloy 718 (ref. 40).

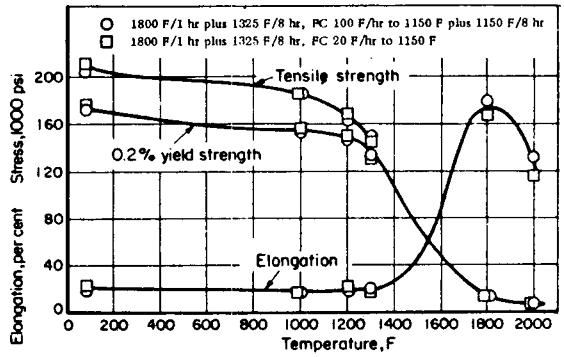

FIGURE 57.—Short-time high-temperature tensile properties—1/2-in.-diam Alloy 718 bar stock.

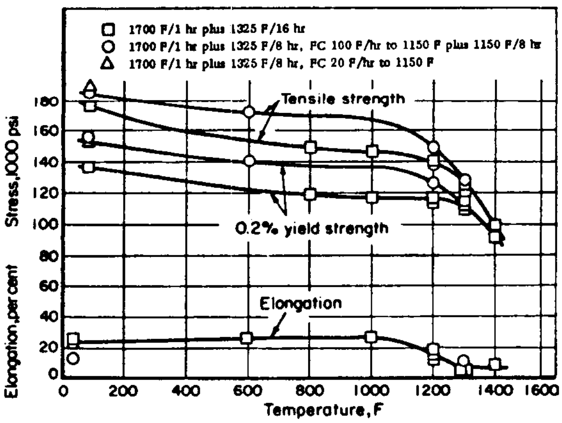

FIGURE 58.—Short-time high-temperature tensile properties—0.060-in.-cold-rolled Alloy 718 sheet.

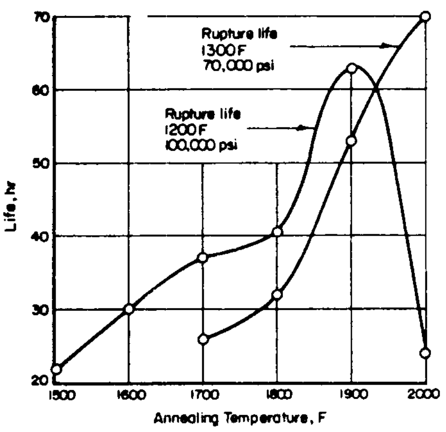

FIGURE 59.—Effect of annealing temperature on rupture life of Alloy 718 (ref. 41). (Annealed 1 hr at temperatures shown and aged at 1325° F for 16 hr.)

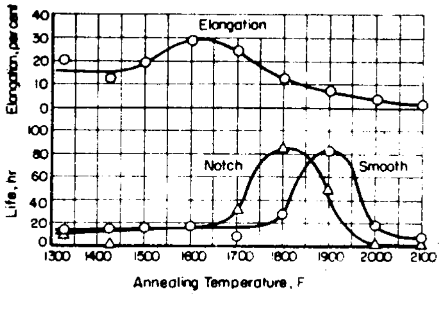

FIGURE 60.—Effect of annealing temperature on stress-rupture properties of Alloy 718 (1300° F/75,000 psi) (ref. 39). [Specimens are from hot-rolled material (1/2-in. diam) annealed for 1 hr, water quenched plus aged at 1325° F/16 hr, AC (0.37% Al, 0.72% Ti, 4.94% Cb plus Ta, and 0.004% B).]

Stress-rupture life is also affected greatly by the annealing temperature. The life at 1300° F and a stress of 75,000 psi was tripled by raising the annealing temperature from 1750° to 1950° F. This is shown by the results in table 57 (ref. 42). Other data showing this effect are presented in figure 59. Figure 60, on the other hand, indicates that maximum creep-rupture life is obtained for notched specimens annealed at 1800° F, rather than 1950° F. Hence, the annealing temperature that provided optimum smooth-bar creep-rupture life produced poor notch-bar life and ductility. Raymond (ref. 37) found that a double-annealing treatment (1950° F/1 hr, FC at 200°/1 hr to 1750° F/1 hr, AC) produced good notch-bar ductility after aging in the 1000° to 1400° F range. This phenomenon was related to the morphology of the carbides and gamma prime near the grain boundaries.

Cullen and Freeman (ref. 52) verified that creep and rupture properties were sensitive to prior thermal history; their findings are summarized in table 58.

When Alloy 718 was annealed at 1950° F, it had high smooth-specimen creep and rupture properties at temperatures up to 1200° F, but very low notched-specimen rupture strength at 1200° F. In contrast to this, material annealed at 1750° F had excellent notched-specimen rupture strength up to 1200° F, but relatively low smooth-specimen creep and rupture strengths. (See table 59.)

The interrelationship between aluminum and titanium content of the alloy, heat treatment, and room-temperature mechanical properties was shown earlier.

The same kind of relationship was found when considering elevated-temperature stress-rupture properties. Some results are summarized in table 60. These indicate that low aluminum, high titanium, and higher annealing temperature result in longer rupture life under the conditions given in the table. However, changes in chemical composition have a greater effect at higher than at lower annealing temperatures. Because of the complexity of these interrelationships the same conclusions may not apply at other temperatures and stresses.

Higher solution treatment and aging temperature were found to increase precracked-Charpy toughness as shown in figure 61. Solution annealing at 1975° F and aging at 1400° F provided optimum properties at test temperatures from −110° to 650° F.

The effect of grain size on the creep-rupture properties of heat-treated Alloy 718 was studied by Stroup and Heacox (ref. 36). They used hot-upset forged sections of a 4.5-in. round-cornered square material that had been rolled from a double-vacuum-melted 20-in.-diam ingot. The forging reductions and temperatures were varied in order to provide material with a variety of grain structures. The forging schedules and stress-rupture properties are summarized in table 61.

Figure 62 shows the beneficial effect of small grains on the creep-rupture ductility at 1200° F. After exposure at 1250° F/50 ksi or 1300° F/50 ksi for

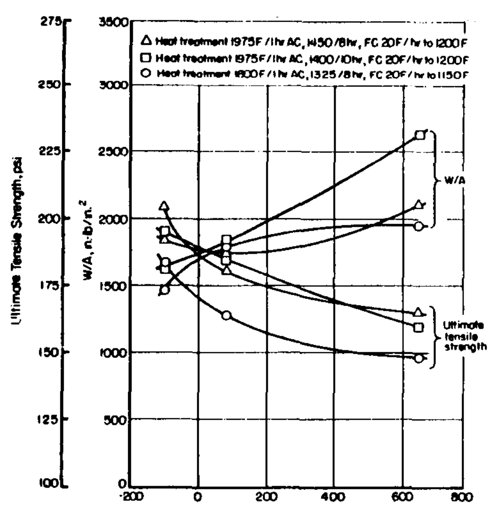

FIGURE 61.–Variation of precracked-Charpy toughness, W/A, and ultimate strength with temperature for three heat treatments of Alloy 718 (ref. 53). (Transverse properties.)

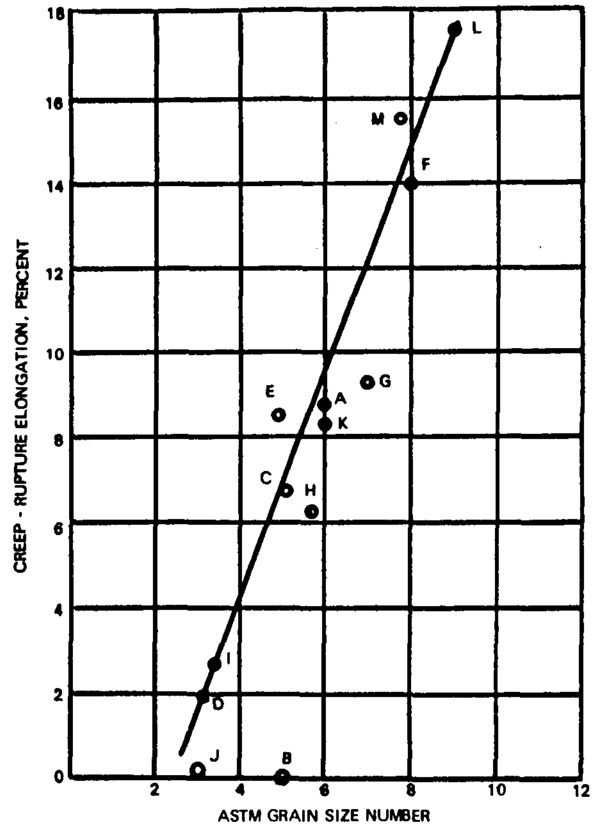

FIGURE 62.–Influence of ASTM grain size on the creep-rupture ductility at 1200° F of upset forged Alloy 718 (ref. 36).

500 hr, rupture ductility increased and the strength decreased due to overaging. The optimum grain size for the best balance of creep-rupture properties was about ASTM no. 6 to no. 7. Both the relative amounts and the size of the grains are important when a duplex grain structure (incomplete recrystallization) exists. Other work has shown similar trends (ref. 54).

The effect of the annealing temperature on the creep-rupture properties of Alloy 718 cast and machined bars is shown in figure 63. Although the minimum creep rate and the rupture life at 1300° F appears better with a 1900° F anneal before aging, the ductility is considerably higher when an 1800° F anneal is used.

Stalker (ref. 55) performed various combinations of cold rolling and heat treatment on 0.625-in.-thick plate that had been "cross rolled" from a 1.4-in.-thick plate. The process history resulted in a total hot reduction of 6:1. The strength increase with greater cold deformation and longer aging at 1200° F is shown

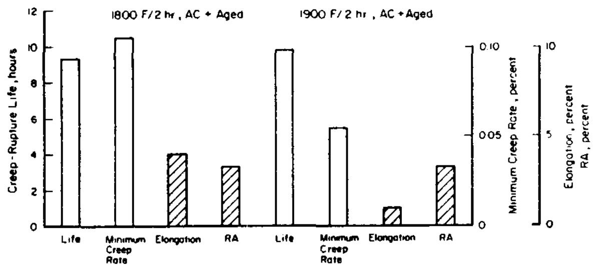

FIGURE 63.–Creep-rupture properties of as-cast Alloy 718 at 1300° F/72,500 psi, showing effect of two annealing temperatures (ref. 45). (Material was vacuum melted and vacuum cast.)

in figures 64, 65, and 66. The ductility dropped as the strength increased.

The benefit of hydrostatic extrusion to both the strength and ductility of wrought Alloy 718 at 1200° F is shown in table 62. The improved ductility is especially noteworthy, in view of the increased strength. Additional data on the high-temperature properties may be found in references 39, 48, and 51. Figure 67 is an illustration of such available data (ref. 51).

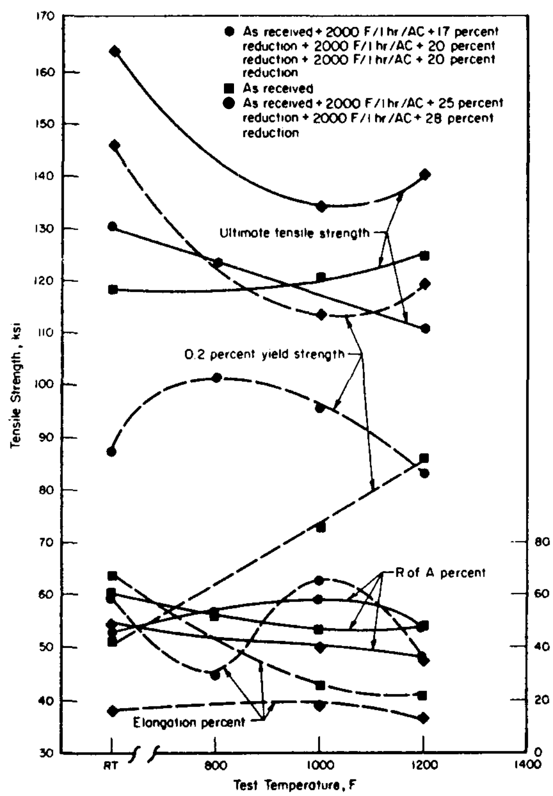

FIGURE 64.–Tensile properties of Alloy 718 plate (ref. 55).

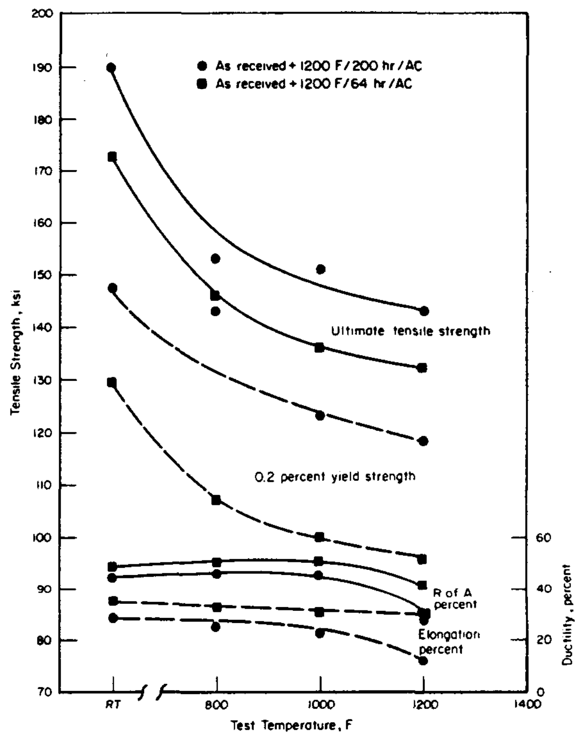

FIGURE 65.–Tensile properties of Alloy 718 plate (ref. 55).

Effects at Cryogenic Temperatures

The alloy exhibits good tensile properties at very low temperatures. A few curves illustrating the effect of various degrees of cold work and of aging are shown in figures 68 to 75. There is a rapid increase in tensile and yield strength as the test temperature decreases to –423° F. Cold working prior to aging results in additional strengthening. Also, lower aging temperature (1250° F) on the 50% cold-reduced sheet results in higher tensile and yield properties and somewhat lower elongation at all test temperatures than is obtained by aging at 1325° F. This is shown in figures 71 to 73. The effects of the same variables on notched tensile strength is illustrated in figures 74 and 75.

The average impact properties of Alloy 718, at three test temperatures, that was given two heat treatments are shown in figure 76. Figures 77 and 78 show that increasing the aging time from 8 to 16 hr improved the cryogenic impact properties of material annealed at 1750° F and at 1800° F, but it decreased the impact strength of material annealed at 1950° F. The effect of additional aging time (20 hr) was differ-

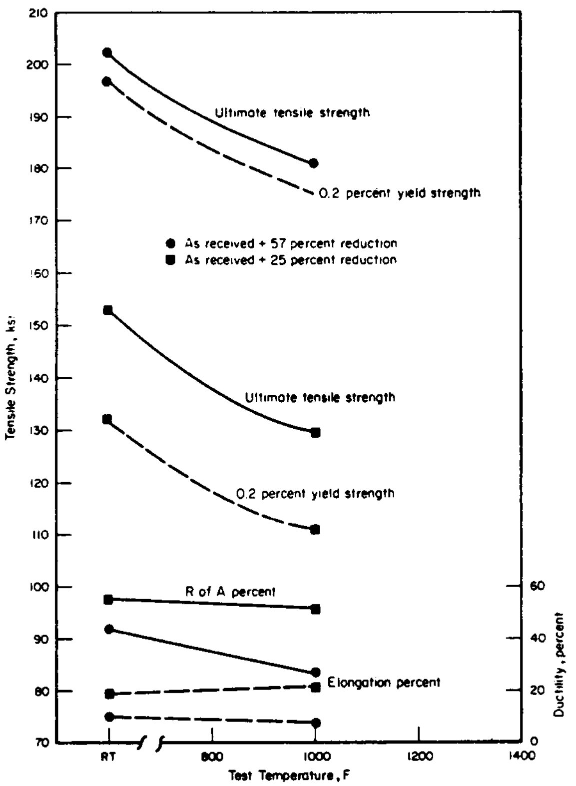

FIGURE 66.–Tensile properties of Alloy 718 plate (ref. 55).

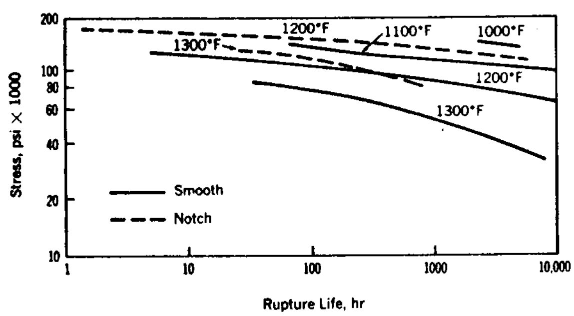

FIGURE 67.–Smooth-bar and notch-bar rupture life of hot-rolled Alloy 718 bar, 5/8-in. diam (1800° F/1 hr, WQ and aged 1325° F/8 hr, FC to 1150° F, hold at 1150° F for total aging time of 18 hr) (ref. 51).

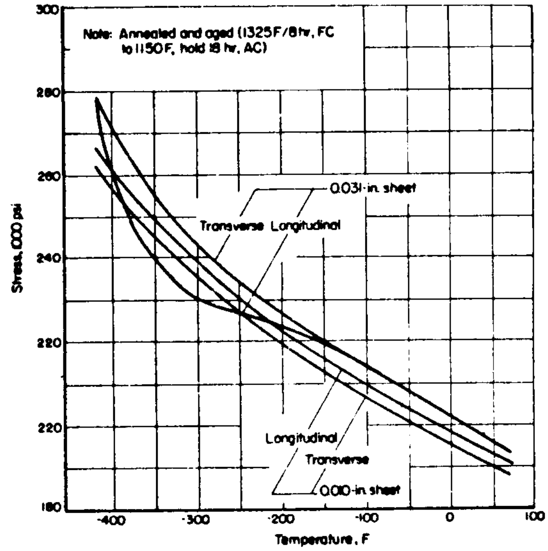

FIGURE 68.–Tensile strength of Alloy 718 at cryogenic temperatures (ref. 56).

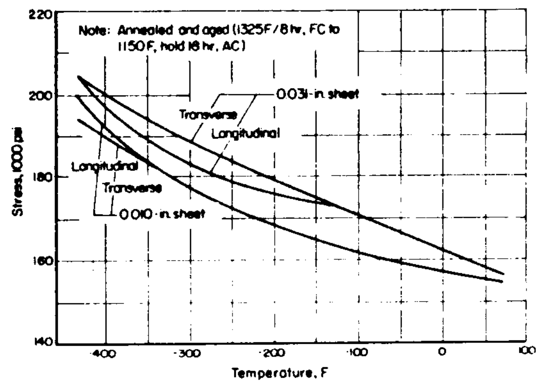

FIGURE 69.–Yield strength of Alloy 718 at cryogenic temperatures (ref. 56).

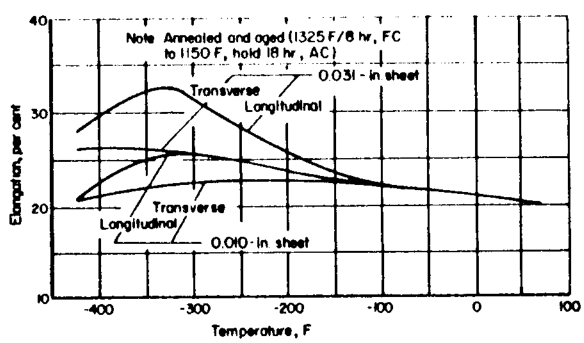

FIGURE 70.–Elongation of Alloy 718 at cryogenic temperatures (ref. 56).

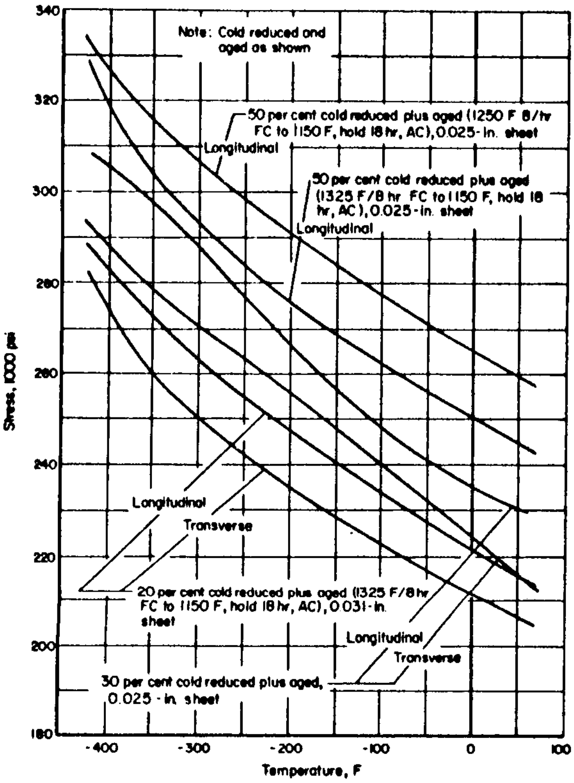

FIGURE 71.–Effect of cold working and aging on the tensile strength of Alloy 718 at cryogenic temperatures (ref. 56).

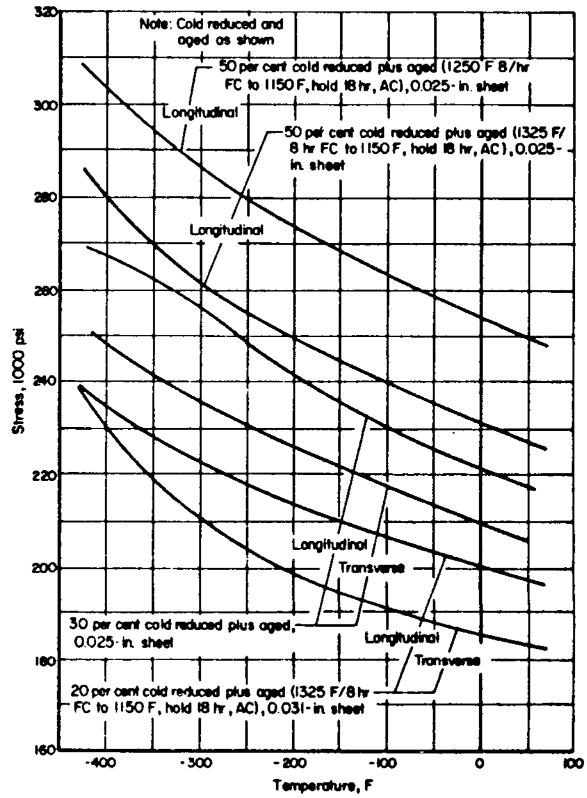

FIGURE 72.–Effect of cold working and aging on the yield strength of Alloy 718 at cryogenic temperatures (ref. 56).

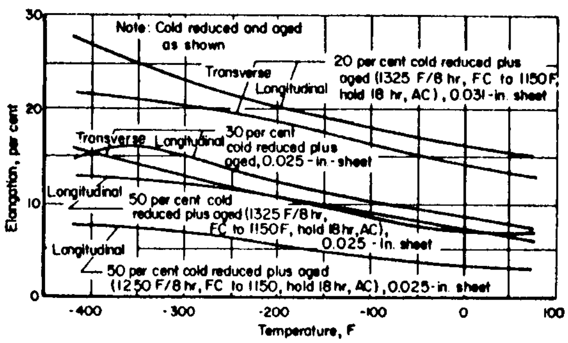

FIGURE 73.–Effect of cold working and aging on the elongation of Alloy 718 at cryogenic temperatures (ref. 56.)

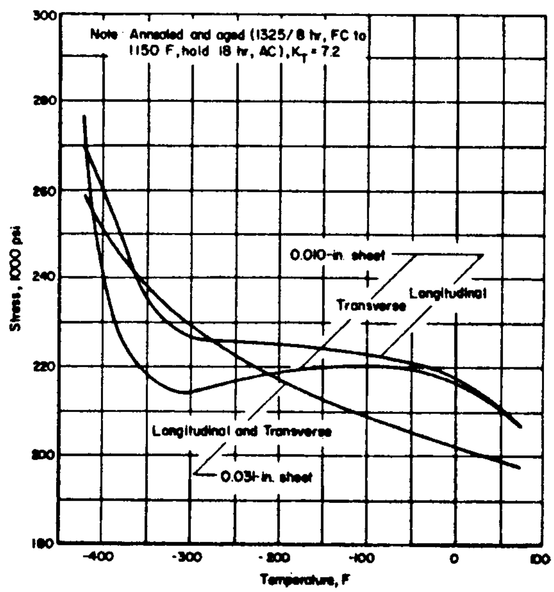

FIGURE 74.–Notch tensile strength of Alloy 718 at cryogenic temperatures (ref. 56).

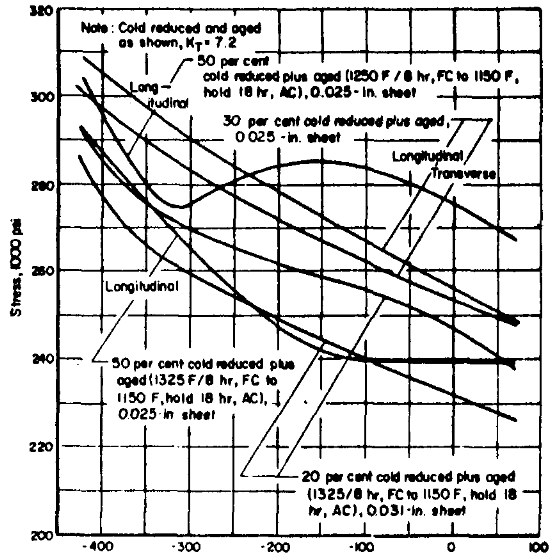

FIGURE 75.–Effect of cold work and aging on the notch tensile strength of Alloy 718 at cryogenic temperatures (ref. 56).

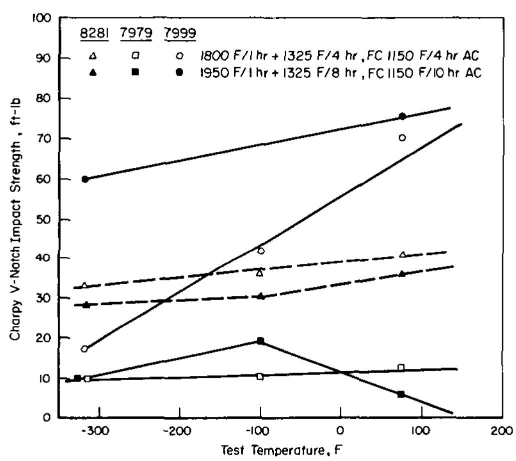

FIGURE 76.–Impact strength as a function of test temperature in three heats of Alloy 718 given two aging treatments (ref. 47).

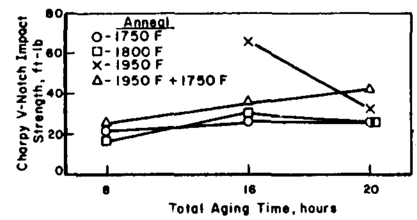

FIGURE 77.–Impact strength of vacuum-melted Alloy 718 at -320° F (heat 7999) showing effect of annealing and aging treatments (ref. 47).

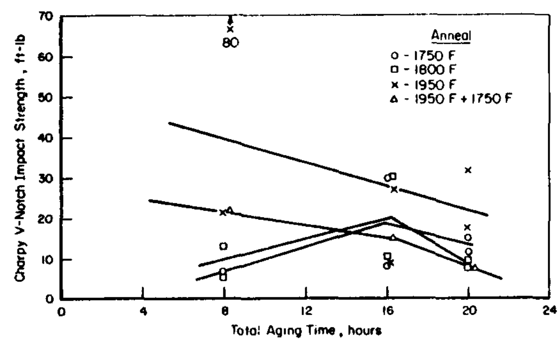

FIGURE 78.–Impact strength of vacuum-melted Alloy 718 at -320° F (heat 7979) showing effect of annealing and aging treatments (ref. 47).

ent in the two heats of Alloy 718 investigated. Improved impact strength with the higher annealing temperature, except in the strongest (most columbium) heat, is shown in figure 79. A definite correlation between higher impact strength and lower columbium content for Alloy 718 is given in figure 80. It is also significant that the heats with lower columbium contained greater amounts of aluminum plus titanium.

Other cryogenic data for Alloy 718 can be found in references 48, 49, 51, 57, 58, and 59.

RENÉ 41

René 41 is a precipitation-hardenable nickel-base alloy possessing high strength in the temperature range of 1200° to 1800° F. The alloy contains cobalt and molybdenum, which provide solid-solution strengthening, but, in common with this class of alloys, high-temperature strength is derived primarily from the precipitation of the $Ni_3(Al,Ti)$ compound dispersed in the matrix. Cobalt also increases the solution temperature of gamma prime and participates in the formation of mu phase. Carbides may also precipitate during aging and affect the mechanical properties. The occurrence of the various phases and their stability and effects on the properties was discussed by Weisenberg and Morris (ref. 60). The relative amounts of gamma prime and the two carbides

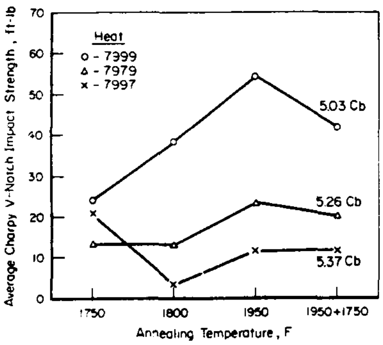

FIGURE 79.–Room-temperature impact strength vs annealing temperature for three vacuum-arc-melted heats of Alloy 718 in billet form (ref. 47).

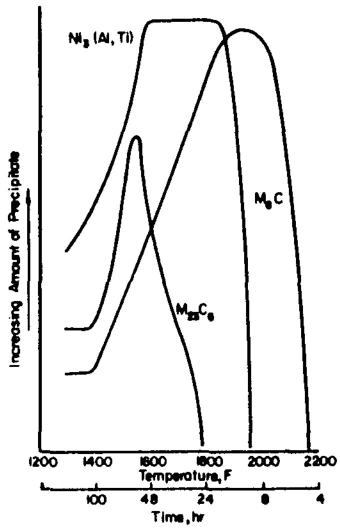

FIGURE 80.–Effect of columbium content on the room-temperature impact strength of vacuum-arc-melted Alloy 718. (Fracture toughness represents several heat treatments tested at room temperature, -100° and -320° F) (ref. 47).

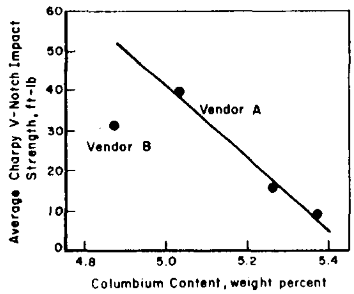

FIGURE 81.–Relative amounts of precipitates in aged René 41 (ref. 60). (Bar stock solution treated at 2200° F, WQ and aged as indicated.)

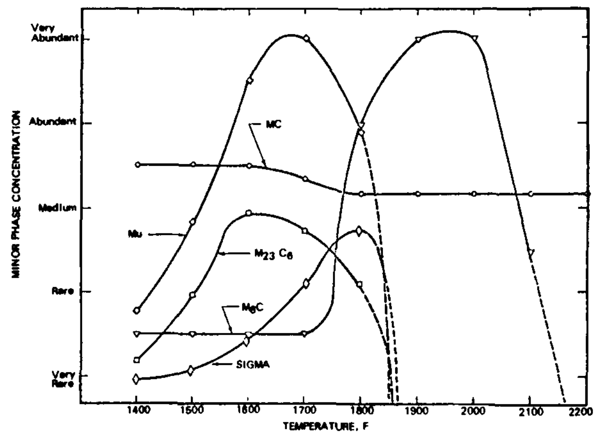

FIGURE 82.–Minor phase concentration in wrought René 41 given 2200° F/2 hr and aged as shown (ref. 61).

that they found after aging at various temperatures are shown in figure 81. The quantity of $Ni_3(Al, Ti)$ precipitated in the matrix is greatest after aging at 1600° to 1800° F, although it may form at as low as 1100° F in cold-worked material. At higher temperatures this precipitate goes back into solid solution, and is completely dissolved at about 1925° F. The behavior and effects of the carbides that are present in the matrix depend on the thermal history of the material. As shown in figure 81, M_6C would not be affected by the ordinary mill-annealing temperature (1975° F) or lower aging temperatures. However, if the alloy is heated to 2150° F, the M_6C would go into solid solution. Any subsequent treatment in the 1400° to 1650° F range will cause the carbon to reprecipitate as $M_{23}C_6$ in a continuous film along the grain boundaries. The more recent work of Collins (ref. 61), in which the Morris heat treatments were used, showed somewhat different results. In this respect, a comparison of figures 81 and 82 shows a difference in the amount of M_6C at 1700° F. Collins also found MC, mu, and sigma phases in addition to M_6C, $M_{23}C_6$, and gamma prime. According to Lacey and Albertin (ref. 62), if the high-temperature annealing is followed by a solution treatment at 1975° F, the carbon reprecipitates as M_6C, also along the grain boundaries. This causes cracking when the alloy is welded and subsequently aged.

Grain size also has an effect on the mechanical properties. Generally the higher solution-treatment temperature (2150° F) promotes grain growth and results in somewhat lower tensile properties, but stress-rupture properties are improved. Conversely, the lower solutioning temperature (1950° F) does not cause grain growth, and optimum tensile properties will be developed on subsequent aging. Stress-rupture properties will be lower because of the smaller grain size.

Various heat-treatments have been devised to provide mechanical properties for practical applications. Examples of the effect of variables in thermal and mechanical treatments on typical properties are given in the following section.

Room-Temperature Properties

The rate of cooling from the annealing or solutioning temperature is very important in the case of René 41 because the rate of gamma-prime precipitation from solid solution is extremely rapid. Therefore, the alloy must be quenched within a matter of seconds in order to retain maximum ductility and softness. The increase in hardness of sheet caused by delay in quenching is shown in figure 83. Because of this, the maximum quench delay for various thicknesses of material was established by The Boeing Co. as given in table 63. A comparison of several quenching procedures is shown in figure 84. Rapid water quenching from the various annealing temperatures results in the best combination of properties (low yield strength and high elongation) required for subsequent forming.

Difficulties with strain-age cracking of weldments prompted an investigation of the aging treatment. A double aging *treatment* was developed that gave increased ductility at elevated temperatures (table 64) with little loss in strength. The effect of various aging treatments on room-temperature yield strength is shown in figure 85 (ref. 63).

The influence of solution-treatment temperature and time on the room-temperature yield and elongation of cold-rolled sheet followed by double aging is shown in figure 86. The lower yield strength and higher elongation obtained by solution treating at high temperatures indicate optimum formability. The curves, however, show that the properties after aging are considerably reduced. In order to retain maximum formability without undue loss in strength after aging, a solution-treatment temperature about 2025° F is indicated.

The tensile properties of René 41 sheet-foil (0.010-in. thick) in the aged and overaged condition is given in table 65 (ref. 64). Other tensile data are readily available from the producers (ref. 65). The large strain-hardening effect of hydrostatic extrusion on annealed and aged wrought René 41 is given in table 66 (ref. 44). Reasonable ductility remains after a 50% cold reduction by hydrostatic extrusion, but

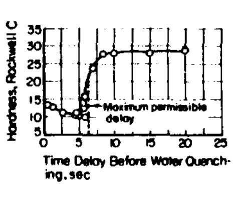

FIGURE 83.–Effect of delay before water quenching on the hardness of 0.070-in. thick René 41 sheet solution heat-treated at 1975° F (ref. 15).

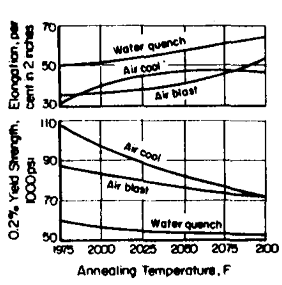

FIGURE 84.–Effect of quenching of René 41 from the annealing temperatures shown (after holding 10 min at temperature) (ref. 62).

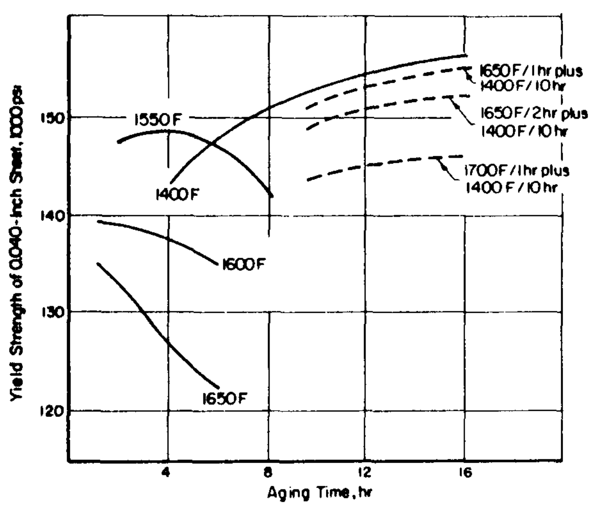

FIGURE 85.–Effect of aging time and temperature on yield strength of René 41.

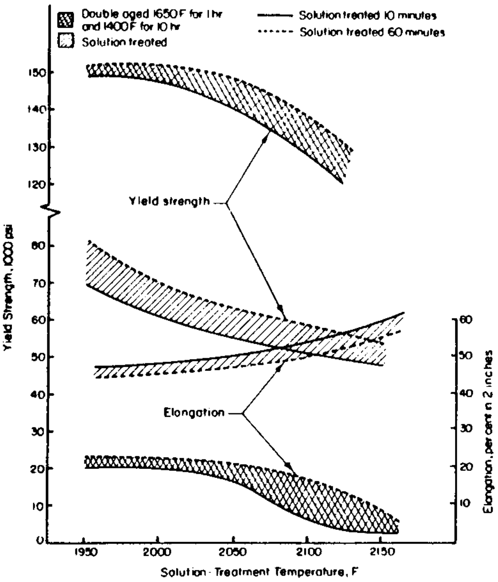

FIGURE 86.—Effect of solution-treating variables on room-temperature yield and elongation of René 41 sheet cold-rolled 20 percent prior to solution treating (ref. 63).

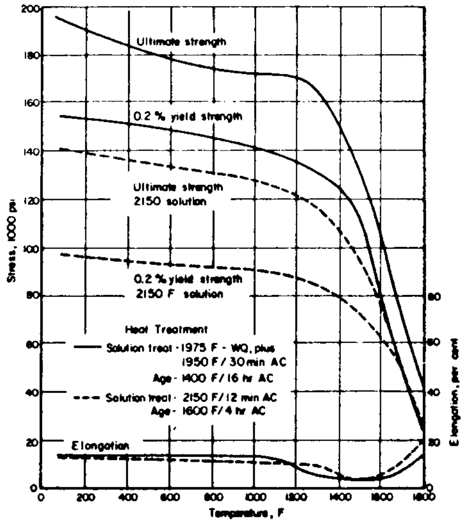

FIGURE 87.—Short-time tensile properties of R-41 sheet as influenced by heat treatment (ref. 7).

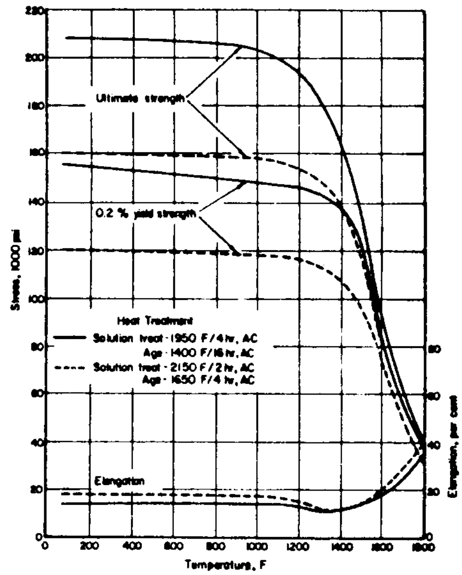

FIGURE 88.—Short-time tensile properties of R-41 bar as influenced by heat treatment (ref. 7).

the material reaches nil ductility (i.e., it is extremely brittle) after post-extrusion aging at 1400° F. Aging at 1400° F/16 hr improved the compressive yield strength and elastic modulus of annealed René 41 cast bars, but it was detrimental to the ultimate compressive strength as shown in table 67 (ref. 45). The impact strength of cast material is presented in table 68 in two conditions of heat treatment (ref. 45). The Mellon Institute studied the tensile, fatigue, and creep-rupture properties of electroslag-melted René 41 processed into several forms and sizes (ref. 66), but as of the date of this writing, a final report has not been issued.

Elevated-Temperature Properties

It was pointed out earlier that optimum tensile properties are obtained when René 41 is solution treated at 1950° F, but that optimum stress-rupture properties are developed following a 2150° F solution treatment. This is illustrated by the short-time tensile data at room and elevated temperature, shown in figures 87 and 88, for sheet and bar, respectively. The comparison for stress-rupture strength is given in table 64. The improvement in ductility at high temperature produced by the double aging procedure is shown in table 69 (ref. 63).

The Abex Corp. (ref. 45) found that when cast test bars of René 41 were heat-treated, the tensile properties were improved. The benefit of the 1950° F/4 hr, AC + 1400° F/16 hr, AC treatment is shown

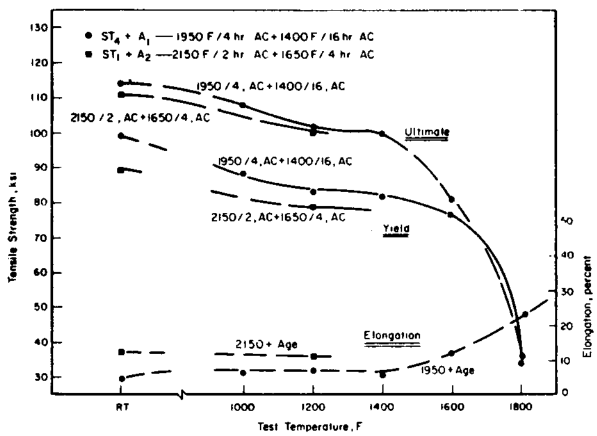

FIGURE 89.–Tensile properties of as-cast René 41 test bars showing the effect of heat treatment.

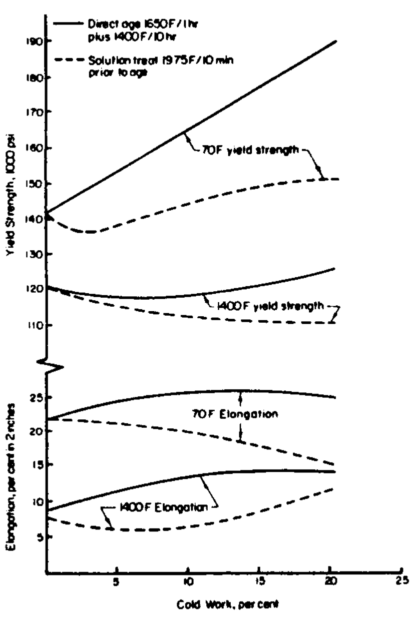

FIGURE 90.–Room and 1400° F yield and elongation of cold-rolled René 41–direct age vs solution treatment and age (ref. 63).

in figure 89 to be greater at the lower test temperatures. A higher ductility corresponds to the lower strength in each instance. Limited creep-rupture tests on cast bars indicated some improvement in properties at 1400° F with the 1950° F/4, AC + 1400° F/16 hr, AC treatment as shown in table 70.

A comparison of room-temperature and 1400° F properties of cold-worked material aged by the double procedure, with and without a prior solution treatment, is given in figure 90 (ref. 45). These results were used to evaluate the need for solution treatment of formed parts, and it was concluded that solution treatment of moderately formed parts was not necessary unless additional forming or welding was to be performed. Additional data showing the effect of amount of cold work on short-time tensile properties over a wide range of temperature are given in table 71 (ref. 31).

Cryogenic-Temperature Properties

Typical properties at cryogenic temperatures are given in table 72 (ref. 26) and figure 91 (ref. 27).

WASPALOY

Waspaloy is also a highly alloyed nickel-base alloy that is hardened primarily by precipitation of $Ni_3(Al, Ti)$ during aging. The effects of heat treatment on the precipitation and solution of gamma prime and $M_{23}C_6$ carbides are similar to those described for René 41. Higher solution-treatment temperatures are used to develop optimum high-temperature creep and rupture strengths, but optimum tensile properties are developed at lower annealing temperatures. The recommended thermal treatments include the solution treatment followed by both a stabilizing and a precipitation-hardening treatment. The difference in short-time tensile properties obtained by these two treatments is shown in table 73. The first treatment, which is recommended for optimum creep and rupture properties, develops somewhat lower short-time tensile strength properties. A comparison of several aging treatments given to annealed and cold-worked material is given in table 74 (ref. 31). Cold working prior to aging results in higher yield and tensile strength and reduced ductility, but relatively shorter

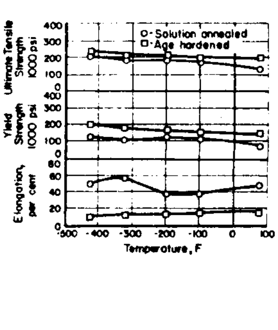

FIGURE 91.–Tensile properties of René 41 at cryogenic temperatures. (0.062-in. sheet, annealed and aged at 1400° F/16 hr.)

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aging times are needed to develop the maximum strength.

The effect of aging treatments on two heats of Waspaloy in four conditions of cold work (between 20% and 40% reduction of area) was studied at Universal-Cyclops, Speciality Steel Div., Cyclops Corp. (ref. 67). The 1/2-in.-diam hot-rolled bars were annealed at 1950° F for 1/2 hr prior to cold drawing. Figure 92 shows the optimum aging temperature to have decreased as either the amount of cold work or the aging time was increased. The maximum hardness was produced by the greatest amount of cold work (40% RA) prior to aging either at 1300° F/8 hr or 1250° F/16 hours. A similar relationship for the room-temperature tensile strength is shown in figure 93 (ref. 67).

The short-time tensile properties, over a range of temperatures from −110° to 1000° F are summarized in table 75 (ref. 31). The effects of stressed exposure at 650° and 1000° F on the properties are also shown. The latter have little effect on the properties of annealed and aged material. However, stressed exposure of sharp-edge notched specimens results in an appreciable decrease in tensile strength. Similar results for creep-rupture and creep properties are shown in tables 76 and 77 (ref. 68).

The properties at cryogenic temperatures are shown in table 78 and figure 94. The ductility of the age-hardened sheet is considerably lower than that of annealed material at the very low temperatures.

Work at Ladish Co. (ref. 75) showed that a combination of programmed thermal cycles and controlled forging reductions brought about greater diffusion of the age-hardening elements. This improvement in homogeneity was deemed responsible for the increase in both transverse tensile strength and

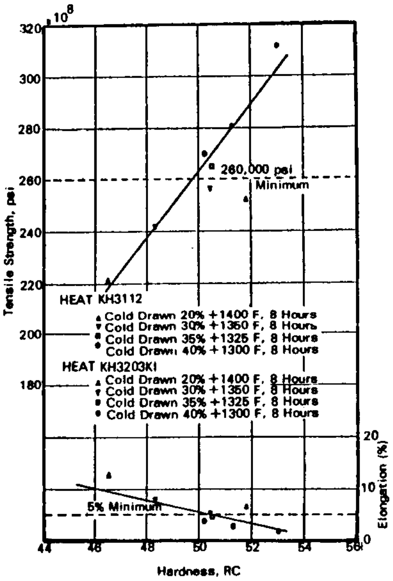


FIGURE 93.—Tensile strength and elongation as a function of hardness for Waspaloy at various conditions of cold drawing and aging.

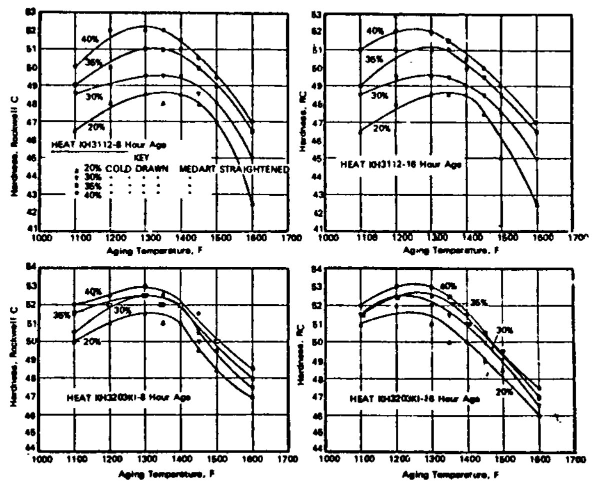


FIGURE 92.—Hardness (Rockwell C) of Waspaloy bar cold-drawn at 20, 30, 35, or 40 percent reduction of area and aged for 8 or 16 hr at the temperature shown (ref. 67).

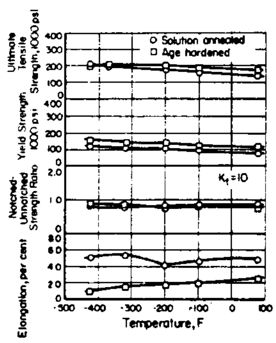


FIGURE 94.—Tensile properties of Waspaloy at cryogenic temperatures (ref. 27). (0.060-in. sheet, annealed and aged 1550° F/2 hr plus 1400° F/16 hr.)

ductility, as shown in table 79. Lower forging temperatures and smaller rates of forging reduction were found to improve forgeability although no details were given.

UDIMET 500

Wrought Form

Udimet 500 is a high-strength nickel-base alloy containing appreciable amounts of cobalt and chromium, and additions of titanium and aluminum to form the age-hardening intermetallic compounds. The alloy is normally double aged to develop the optimum properties (ref. 15). Solution treatment at 1950° to 2150° F is followed by air cooling. All phases except TiC are dissolved by the treatment, but considerable $Ni_3(Al, Ti)$ is precipitated during cooling. The first aging step is at 1550° F, which precipitates $M_{23}C_6$ in a discontinuous form at the grain boundaries. This increases creep-rupture life but tends to decrease ductility. Final aging, 1400° F, determines the ultimate hardness of the alloy by controlling the size of the $Ni_3(Al, Ti)$ precipitate. Higher room-temperature ductility can be obtained by omitting the 1550° F treatment, but subsequent service temperatures in the 1500° to 1600° F range will cause precipitation of large amounts of $M_{23}C_6$ in the grain boundaries, again decreasing the ductility.

Two standard heat-treatment schedules have been developed, depending on the end properties desired. The first, for optimum tensile properties, includes a solution treatment at 1975° F for 4 hr, and air cooling, followed by the two aging steps mentioned above. The other, suggested for best high-temperature creep and rupture strength, includes a preliminary 2-hr treatment at 2150° F, and air cooling, followed by the same 1975°, 1550°, and 1400° F steps as in the first schedule.

The room- and elevated-temperature tensile properties obtained by the first treatment are given in table 80, while table 81 and figure 95 give the stress-rupture properties obtained by treating according to the procedure recommended for optimum high-temperature rupture properties.

The effect of omitting the 1550° F treatment to improve room-temperature tensile strength and ductility was mentioned above. Data supporting this statement are given in table 82.

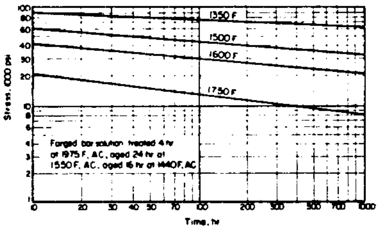


FIGURE 95.–Stress-rupture strength of forged Udimet alloy 500 bar (ref. 6). (Heat treatment: 2150° F/time–dependent on section size–rapid air cool; 1975° F/4 hr, AC; 1550° F/24 hr, AC; 1400° F/16 hr, AC.)

Cast Form

The alloy is also produced in the form of vacuum-investment castings. Castings are supplied in the as-cast condition, but the following heat-treatment schedule has been found beneficial (ref. 6):

1. Solution heat treat 4 hr at 2100° F, AC
2. Solution heat treat 4 hr at 1975° F, AC
3. Age 16 hr at 1400° F, AC.

HASTELLOY C

Hastelloy C is a nickel-base alloy noted for its good resistance to aggressive chemicals and to oxidizing and reducing gaseous atmospheres. It retains useful mechanical properties at elevated temperatures up to about 1600° F. The alloy contains chromium, molybdenum, iron, and tungsten in amounts almost to the extent of their mutual solubility (ref. 69). In contrast to the other precipitation-hardenable alloys considered in this report, Hastelloy C derives its strength almost entirely from solid-solution hardening. Aluminum and titanium, which form the $Ni_3(Al, Ti)$ hardening compound, are essentially absent in this alloy. However, a certain amount of hardening can occur due to the precipitation of Ni_7Mo_6 and M_6C.

The alloy is supplied in the solution-heat-treated condition. Wrought products are solution heat-treated at 2225° F and water quenched. The Ni_7Mo_6 and some M_6C particles are dissolved at this temperature. The time at temperature depends on the section thickness, but quenching must be rapid to prevent precipitation of the secondary phases Ni_7Mo_6 and M_6C as a harmful film along the grain boundaries.

Sand castings are solution annealed at 2250° F, and rapid air cooled (RAC) or water quenched. Overheating of the alloy by as little as 50° F in excess of these values may impair its tensile strength (ref. 4). Investment castings are usually supplied in the as-cast condition, but they may be aged to increase room-temperature tensile strength at the expense of ductility. The effect of variables in the thermal and mechanical treatments on the mechanical properties of the alloy are discussed in the following sections.

Room-Temperature Properties

Table 83 (ref. 4) shows typical hardness values for various forms of Hastelloy C and the influence of aging on the solution-treated sheet and investment-cast specimens. It appears that the aging effect is relatively minor on the sheet specimens, but is somewhat more pronounced on investment castings. Room-temperature tensile data are summarized in table 84 (ref. 4). The aging conditions shown increase the tensile strength slightly, but elongation is decreased appreciably by heating for longer periods and at the higher temperature. Long-time aging of investment castings reduces the elongation practically to zero.

Reducing the solution-treatment temperature has been reported by the Stellite Works* (ref. 69) to have a strengthening effect on the alloy with some reduction in ductility. These effects are shown in figure 96. Similar results were reported by Grossman (ref. 70). Experiments were made to evaluate a 2050° F solution treatment in comparison with the recommended 2225° F. The results on tensile properties at room temperature and at 1400° F are summarized in table 85. The slight increase in ultimate strength, decrease in yield strength, and definite reduction in elongation were attributed to a slight age-hardening effect produced by the incomplete solid solubility at the lower temperature. If these properties are acceptable, the lower annealing temperature would make available a number of heat-treating furnaces that cannot operate above 2100° F.

The effect of cold working on the tensile properties is indicated in table 86. As expected, cold working increases the tensile and yield strengths, while decreasing the ductility.

*Union Carbide Corp.

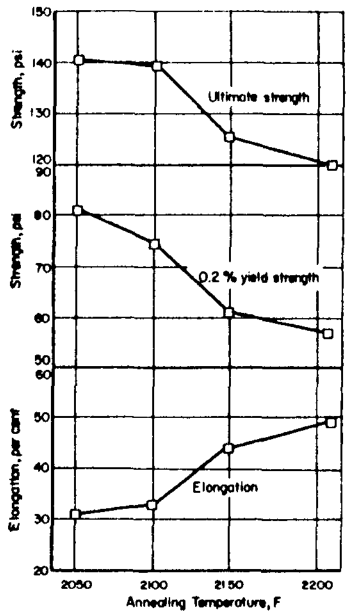


FIGURE 96.–Effect of final annealing temperature on the tensile properties of 0.050-in. thick Hastelloy C sheet.

High-Temperature Properties

Typical high-temperature tensile properties of sheet and cast Hastelloy C are given in tables 87 and 88. Stress-rupture data are shown in table 89 (ref. 4).

These tables give only a limited amount of data on the effect of changes in thermal and mechanical treatments on the properties because Hastelloy C is rather insensitive to changes in heat treatment, and the conditions given are those recommended for optimum properties.

ALLOYS 713C and 713LC

Alloys 713C and 713LC are precipitation-hardenable, vacuum-melted, vacuum-cast nickel-base alloys having excellent strength properties up to 1800° F. Alloy 713LC is a low-carbon modification (0.05 C) of Alloy 713C (0.12 C). It was developed to obtain increased room-temperature tensile properties while maintaining the good stress-rupture properties of the base alloy 713C. The effect of carbon content on these properties is shown in figures 97 and 98 (ref. 12). The best combination of good tensile

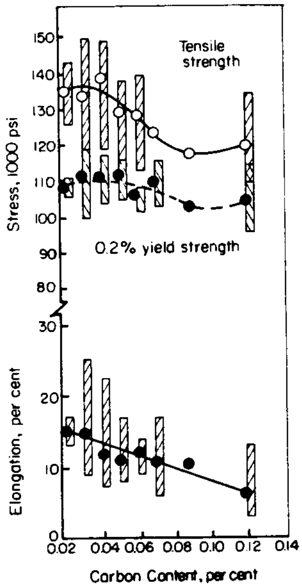


FIGURE 97.–Room-temperature tensile properties vs carbon content of as-cast, vacuum-melted, vacuum-cast Alloy 713LC.

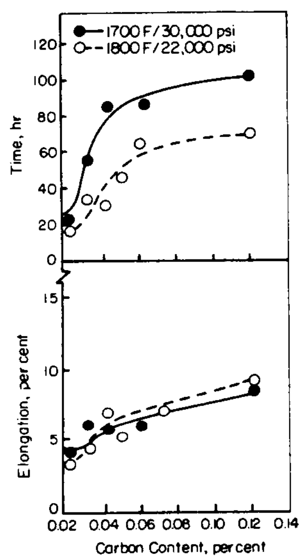


FIGURE 98.–Creep-rupture properties vs carbon content of as-cast, vacuum-melted, vacuum-cast Alloy 713LC.

properties at room temperature and stress-rupture properties at high temperature is obtained in the 0.03 to 0.07 carbon range.

Alloy 713C is normally used as-cast, and typical short-time tensile properties are shown in table 90 (ref. 11). Stress-rupture properties are given in table 91 (ref. 11). Heat treatment of the alloy has been investigated. Heating for 2 hr at 2150° F followed by air cooling increased the rupture life at 1700° F appreciably, with a reduction in ductility from 7% to 4% in elongation. In this condition, however, the alloy suffered a sharp decrease in rupture strength and ductility at 1350° F. Further investigation showed that stabilizing at 1700° F, following the 2150° F solution treatment, was required to retain optimum properties at 1359° F. These results are summarized in table 92 (ref. 11).

Some data have been accumulated on the impact properties of Alloy 713C as related to grain size and heat treatment; these are shown in figure 99. In general, cast alloys do not attain the high level of impact strength shown by wrought alloys. Solution treatment of the alloy offers a slight improvement in impact strength up to about 1300° F.

The fatigue strength of Alloy 713C in several conditions has been determined. Some of the data, replotted in Alloy Digest, D-1128 (ref. 71), are shown in figures 100 to 103. The effects of a 2150° F treatment followed by aging for 2 hr at 1700° F, compared with as-cast results, are shown in figures 100 and 101. Several treatments, representative of conditions encountered in brazing, followed by stress relieving or

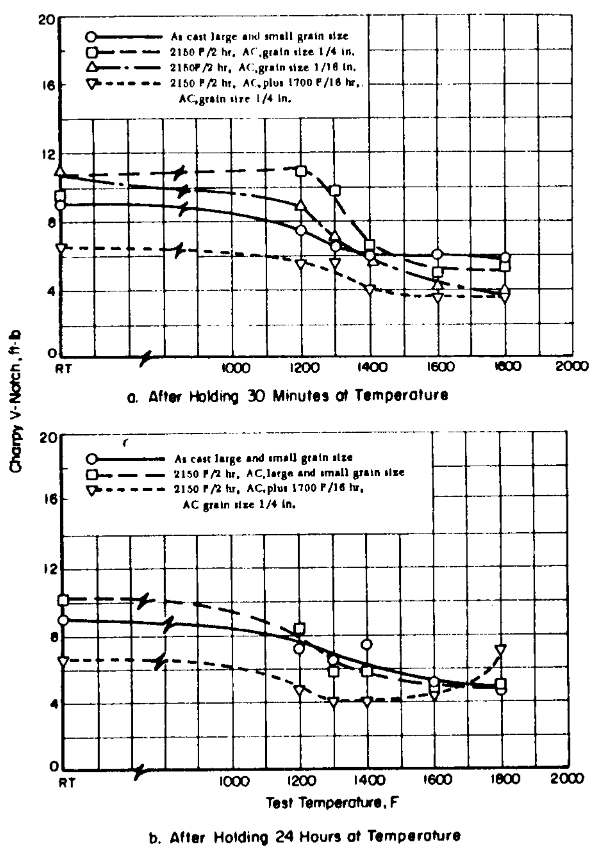


FIGURE 99.–Impact properties of as-cast, vacuum-melted, vacuum-cast Alloy 713C (ref. 11).

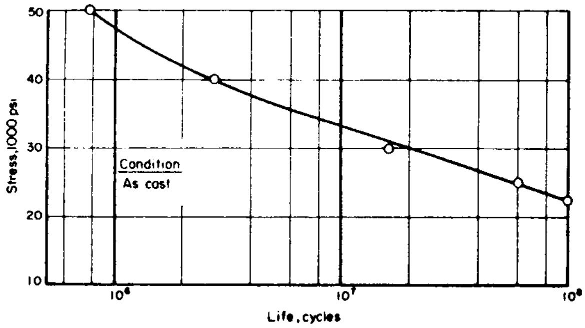


FIGURE 100.–R. R. Moore fatigue strength at room-temperature of Alloy 713C (ref. 71).

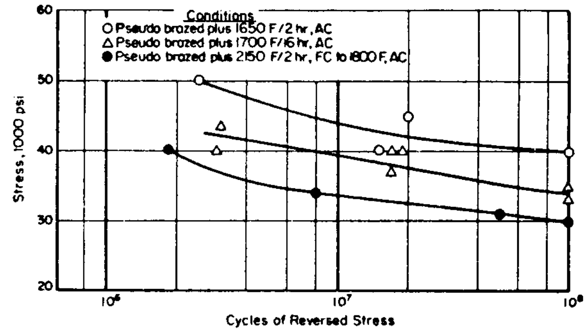


FIGURE 102.–Rotating-beam fatigue strength of Alloy 713C (single vacuum-cast, coarse-grain) related to heat treatment (ref. 71).

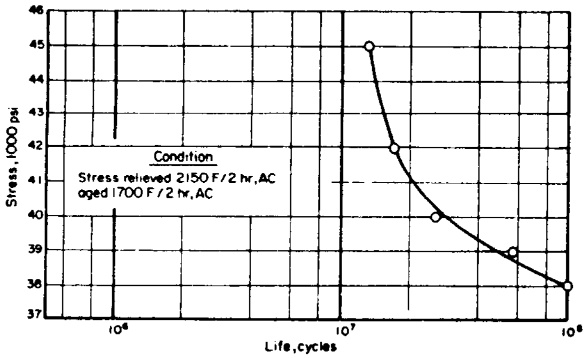


FIGURE 101.–R. R. Moore fatigue strength at room temperature of Alloy 713C (ref. 71).

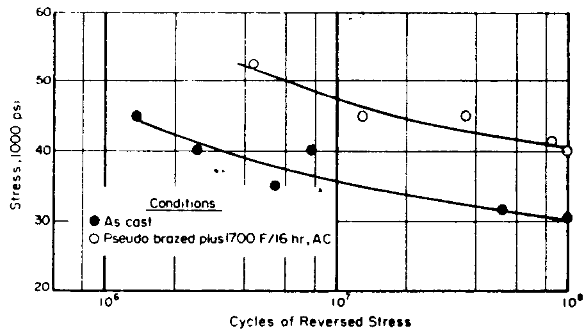


FIGURE 103.–Rotating-beam fatigue strength of Alloy 713C (single vacuum-cast, fine-grain) related to heat treatment (ref. 71).

aging, are compared in figures 102 and 103. These results all indicate that fatigue life is definitely affected by the prior thermal history of the material.

Very little additional data are available on the effect of heat treatments on the properties of Alloy 713LC. One comparison, showing the influence of a solution treatment on the room-temperature tensile properties of low-iron and high-iron 713LC is made in table 93. Apparently, solution treatment improves the tensile and yield properties in both materials. However, the ductility of the low-iron material is decreased, whereas little effect is evident on the high-iron material.

Increasing iron content has been shown to decrease the high-temperature stress-rupture life of Alloy 713LC (ref. 12).

IN-100 ALLOY

IN-100 is a nickel-base casting alloy developed for use at temperatures up to 1900° F. As shown in table 1, the alloy contains chromium, cobalt, molybdenum, and relatively large amounts of aluminum and titanium. The alloy is strengthened by precipitation hardening (ref. 72), the $Ni_3(Al,Ti)$ gamma prime precipitating from the matrix on cooling of the casting from the solidification temperature; or on cooling after solution treatment. IN-100 also contains a large amount of MC (M being predominantly titanium). $M_{23}C_6$ can form during elevated-temperature exposure as MC breaks down.

A study was made to determine the optimum solution-treatment temperature (ref. 73) by metallographic examination of specimens heat-treated for 2 hr over a 2100° to 2200° F temperature range in argon. It was concluded that complete solution of the gamma prime occurs only between 2170° and 2200° F. Eutectic melting was observed at 2200° F, so solution treating at this temperature is not advisable. Only partial solution of the gamma prime takes place in the 2100° to 2170° F temperature range.

The effect of solution treatments at 1900°, 2050°, and 2150° F on tensile and rupture properties were also examined in this investigation. The results are presented in tables 94 and 95, respectively. All three treatments reduced the tensile properties at 1300° F from those in the as-cast condition. The rupture results show that the life at 1800° F/29,000 psi decreased gradually from an average of 33 hr in the as-cast condition to approximately 28 hr in the 1900° F solution-treated condition.

A comparison of the short-time tensile properties of the alloy as-cast and solution-treated is shown in figures 104 to 109. These indicate the same general effects reported in tables 94 and 95 from another investigation. The solution treatment reduces the tensile properties over the entire range of test temperatures. However, there is appreciable scatter in some of the results obtained from replicate test, particularly in elongation. More testing is probably needed to establish the usable property ranges.

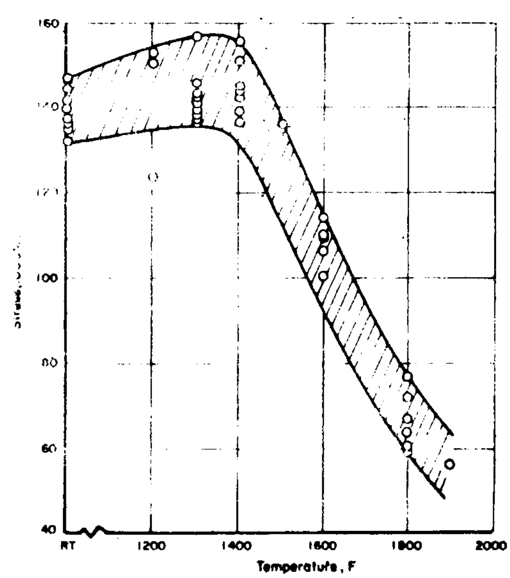


FIGURE 104.–Relation of ultimate tensile strength to temperature for as-cast IN-100 (ref. 71).

MAR-M 200 ALLOY

MAR-M 200 is a nickel-base precipitation-hardening alloy designed primarily for cast turbine blades that operate at service temperatures up to 1900° F. It contains a relatively large amount of tungsten and some columbium for solid-solution strengthening,

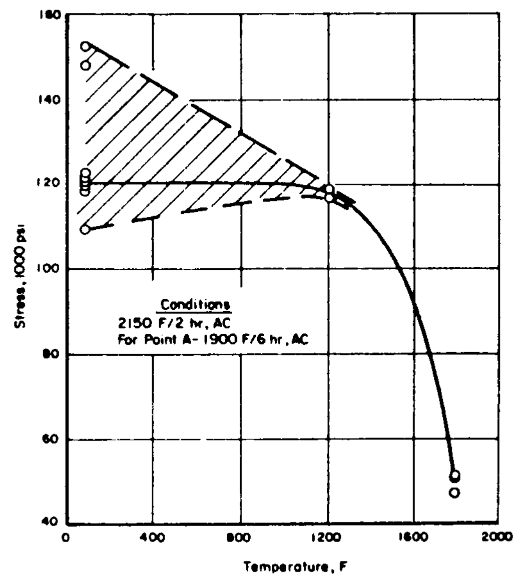


FIGURE 105.–Relation of ultimate tensile strength to temperature for heat-treated IN-100 (ref. 71).

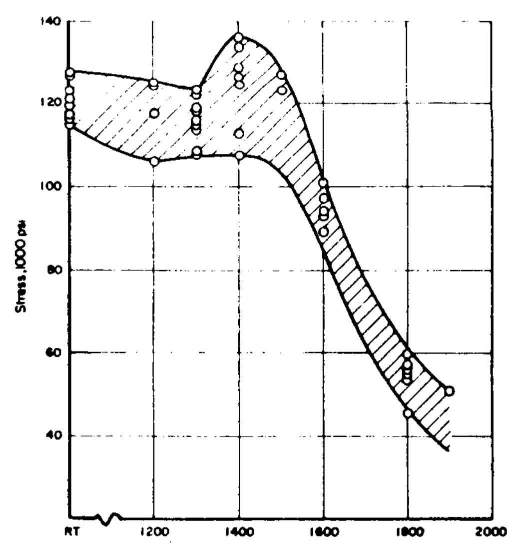


FIGURE 106.–Relation of 0.2 percent yield strength to temperature for as-cast IN-100 (ref. 71).

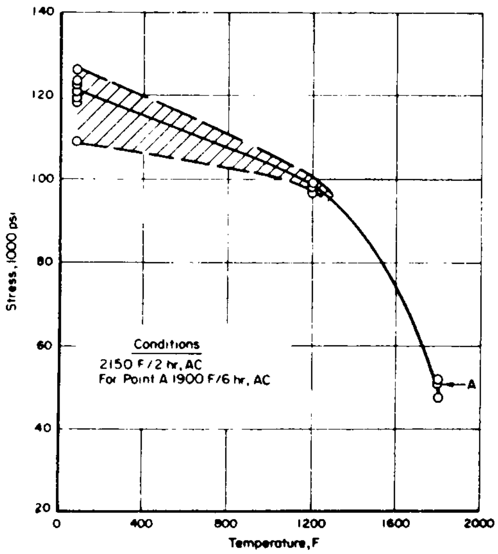


FIGURE 107.—Relation of 0.2 percent yield strength to temperature for heat-treated IN-100 (ref. 71).

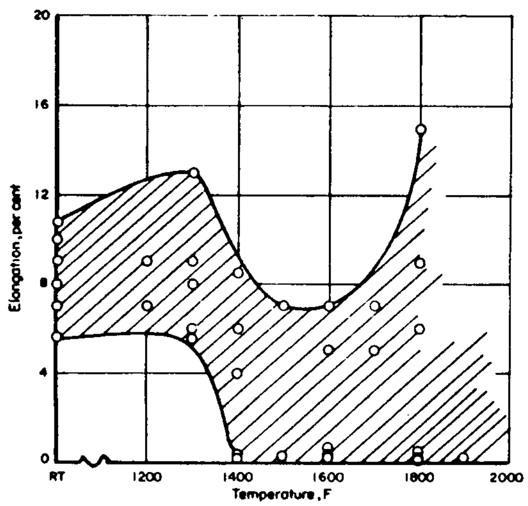


FIGURE 108.—Relation of elongation to temperature for as-cast IN-100 (ref. 71).

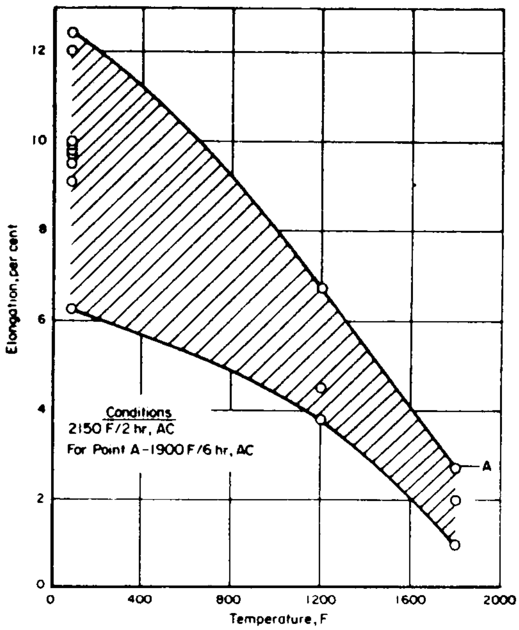


FIGURE 109.—Relation of elongation to temperature for heat-treated IN-100 (ref. 71).

and normal amounts of aluminum and titanium for precipitation hardening. Cobalt is added to increase the solution temperature of gamma prime, and therefore the softening temperature of the alloy. MAR-M 200 has only moderate resistance to oxidation because of its relatively low chromium content. This alloy has good casting characteristics and is melted and cast in vacuum in either shell or investment molds.

Cast MAR-M 200 has a microstructure consisting of globular (interdendritic) and intergranular (eutectic) gamma prime as well as massive MC carbides in the grain boundaries. The eutectic gamma-prime phase dissolves between 2100° and 2200° F, and tends to form a continuous intergranular film at 2000° F and below. Both $M_{23}C_6$ and M_6C phases precipitate at 2000° F and below. Table 96 shows the phases appearing after certain aging treatments.

One investigation indicated that a 2000° F/4-hr treatment improved rupture life at 1400° F and tensile strength at room temperature, but reduced ductility (ref. 74).

Creep tests at 1800° F (fig. 110) showed that precipitation of $M_{23}C_6$ and gamma-prime agglomeration had no significant effect on the creep rate of MAR-M 200, but M_6C precipitation caused weakening. A comparison of curves 2 and 3 in figure 110 shows that high-temperature weakening was caused by M_6C precipitation. The fine M_6C particles created at 2000° F changed to platelets during rupture testing at 1800° F.

Figure 111 shows that improvement of the average stress-rupture strength at 1400° F was obtained with a heat treatment of 1650° F/16 hr + 1600° F/40 hours (ref. 75). Precipitation of finer gamma-prime particles during heat treatment is thought to be partly responsible for the improved strength. No improvement in rupture strength was noted at 1800° F.

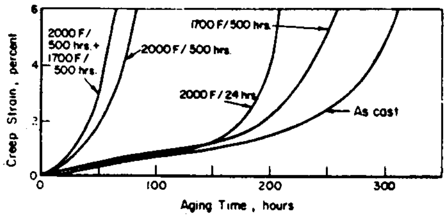


FIGURE 110.–Creep curves at 1800° F for directionally solidified MAR-M 200 at 25 ksi stress showing the effect of heat treatment on strain rate and time to third-stage creep (ref. 77).

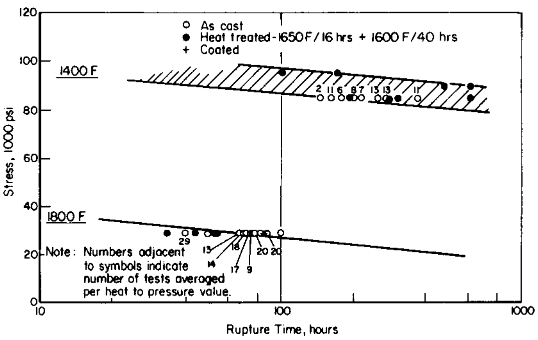


FIGURE 111.–Stress-rupture curves for MAR-M 200 alloy at 1400° and 1800° F (ref. 75). (Heat treatment showed improved strength in some instances, but not all.)

Cast jet-engine turbine blades have been made by directional-solidification and single-crystal methods because of the improved strength and ductility so obtained. Since propagation of intergranular cracks oriented normal to the stress axis often causes service failure of gas-turbine blades, the elimination of these normal or transverse grain boundaries would seem beneficial. Pratt & Whitney conducted engine tests, at 1900° and 2000° F, of blades and vanes, respectively, which demonstrated improved life and thermal resistance of directionally solidified and single-crystal parts over conventional castings of MAR-M 200 (ref. 76).

Heat treatment (2250° F/4 hr + 1600° F/64 hr) of directionally solidified MAR-M 200 (designated DS-200) improved its yield strength significantly over the 75° to 1500° F temperature range as shown in figure 112 (ref. 77).

Slow cooling of cast DS-200 below about 2100° F results in "overaging" and causes the strength to drop;

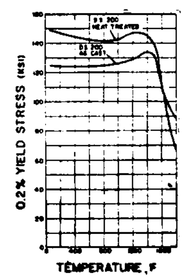
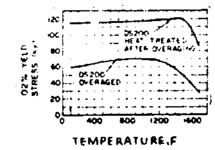


FIGURE 112.–Yield strength vs temperature for directionally solidified MAR-M 200 (DS 200) heat-treated at 2250° F/4 hr + 1600° F/64 hours.

FIGURE 113.–Yield strength vs temperature for directionally solidified MAR-M 200.



the microstructure of the overaged material has a dispersion of coarse gamma-prime particles. However, figure 113 (ref. 77) shows that the strength of the casting can be restored by a resolution and aging treatment.

Heat treating the as-cast DS-200 material at 2250° F/4 hr + 1600° F/64 hr significantly reduced the creep rate at 1800° F, but had only a slight effect at 1400° F, as shown in figure 114 (ref. 77).

Piearcey et al. (ref. 78) found that single crystals* oriented in single-slip directions exhibit maximum ductility, minimum work hardening, and the lowest creep resistance; whereas crystals in multiple slip orientations show minimum ductility, maximum work hardening, and the greatest resistance to creep. The creep strength was much less dependent on crystal orientation as the test temperatures increased from 1400° to 1800° F.

As table 97 shows, heat treatment (2250° F/1 hr + 1600° F/32 hr) also had an important effect on the rupture life at 1400° F; the heat-treated material was considerable stronger in each of the orientations shown.

The effect of solution treatment with subsequent air cooling on the room-temperature tensile behavior of cube-oriented** MAR-M 200 crystals, 2300° F appearing to be the optimum temperature, is shown in figure 115 (ref. 78). A loss in ductility accompanied the improved yield strengths; the tensile strength of the material treated at 2300° F was not substantially

*Grown under vacuum by a modified Bridgman method, i.e., gradient cooling.

**The [100] tensile axis coincided with the growth axis.

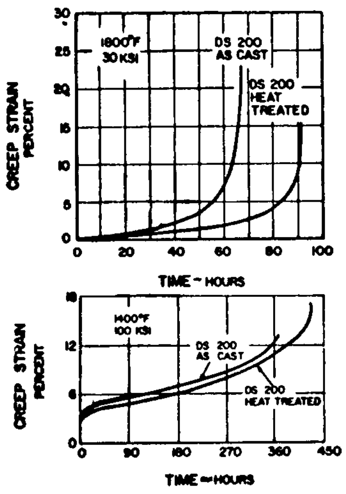


FIGURE 114.–Creep properties of directionally solidified MAR-M 200 in the as-cast and the heat-treated conditions.

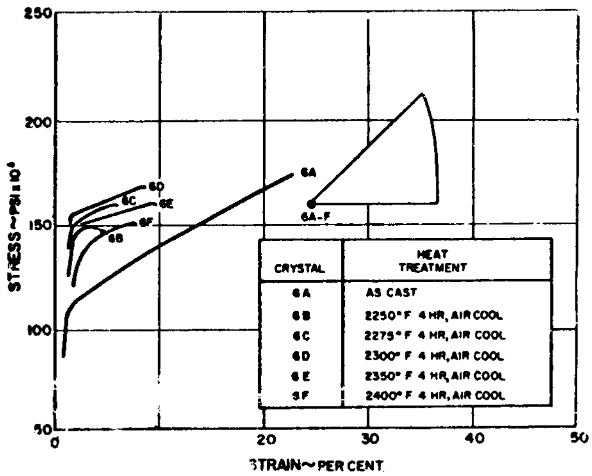


FIGURE 115.–Effect of solution temperature on the 70° F tensile behavior of cube-oriented MAR-M 200 single crystals.

lower than the as-cast material, but the tensile strength of the other material was lowered.

Higher yield strengths of heat-treated material (2250° F/1 hr, AC + 1600° F/32 hr) are shown in figure 116 (ref. 78) at four different test temperatures. It is noteworthy that the yield strengths are similar for solution-treated and air-cooled material (~157 ksi) (fig. 115) and for solution-treated and aged material (~165 ksi). This indicates that hardening occurred because of gamma-prime precipitation during cooling.

TAZ-8 ALLOYS

Research at Lewis Research Center has resulted in the development of a high-strength cast nickel-base alloy series designated TAZ-8 (refs. 79 to 84). The first alloy of this group (TAZ-8) has the following composition (percent):

Ni: balance (68)	W: 4
Cr: 6	Zr: 1
Ta: 8	C: 0.125
Al: 6	V: 2.5
Mo: 4	

This alloy compares favorably to the stronger commercial alloys; i.e., it has a tensile strength exceeding 50,000 psi at 1900° F in the as-cast condition (ref. 85). Although it is basically a cast alloy, thickness reductions of 50% were obtained with 1/2-in. diam as-cast bars by unidirectional forging techniques at room temperature. Cast sheets, 0.100-in. thick, have been rolled to 0.015 in. by special techniques.

Not much has been published about the microstructural constituents in the TAZ-8 series of alloys, but work by Collins (ref. 86) with similar alloys has shown three types of carbides and a boride to exist in the as-cast condition. From these data and data on other commercial alloys (refs. 87 and 88) the following phases may be anticipated in TAZ-8 alloys in one condition of heat treatment or another:

Compound	*Main Constituents*
MC	M = Ta + (Cb*)
MC′	M = Zr + Ta + (Cb*)
M_6C	M = Mo + W
M_3B_2	M = Mo
MN	M = Zr
$M_{23}C_6$	M = Cr
γ′ (Gamma prime), Ni_3 (Al, Ti) age-hardening phase	
A_7B_6–Mu phase, Ni $(Mo, W)_6$ after very long exposure times only	

*Present in TAZ-8A alloy but not TAZ-8.

The TAZ-8 alloys have very high recrystallization temperatures, which accounts for part of their resistance to softening and creep at 1900° F. Material

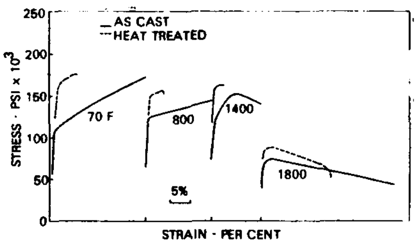


FIGURE 116.–Effect of heat treatment (2250° F/1 hr + 1600° F/32 hr) on the tensile behavior of cube-oriented MAR-M 200 single crystals.

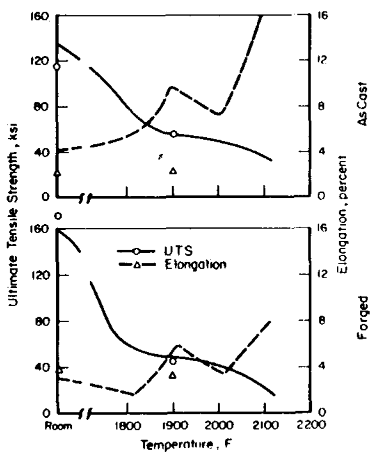


FIGURE 117.–Tensile properties of NASA TAZ-8 alloy in the as-cast and the forged conditions. [The isolated points (at RT and at 1900° F) are from material with 1.8% V instead of the usual 2.5%.]

cold-worked 14% and annealed at 2150° F/½ hr did not recrystallize. TAZ-8 partially recrystallized after hot rolling 45% and annealing for 1/2 hr at 2200° F. Complete recrystallization was accomplished by hot working 60% at 1650° F and then annealing at 2200° F/½ hour.

Annealing between 2000° and 2100° F dissolves some, but not all, of the as-cast precipitates. Since the solidus temperature of the alloy is near 2300° F, complete solution of the second phases appears unlikely.

The tensile strength of TAZ-8 alloy (with 2-1/2% V) at test temperatures from 1800° to 2100° F is greater in the as-cast condition than after forging, as illustrated in figure 117 (ref. 89) and table 98 (ref. 90).

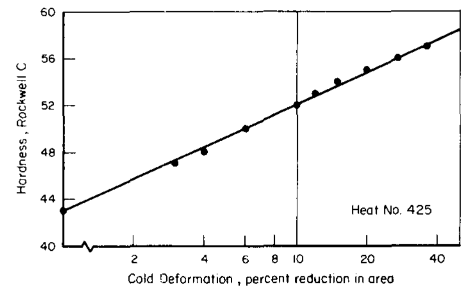


FIGURE 118.–The work hardening of high-carbon TAZ-8 alloy is shown above by the hardness vs deformation plot.

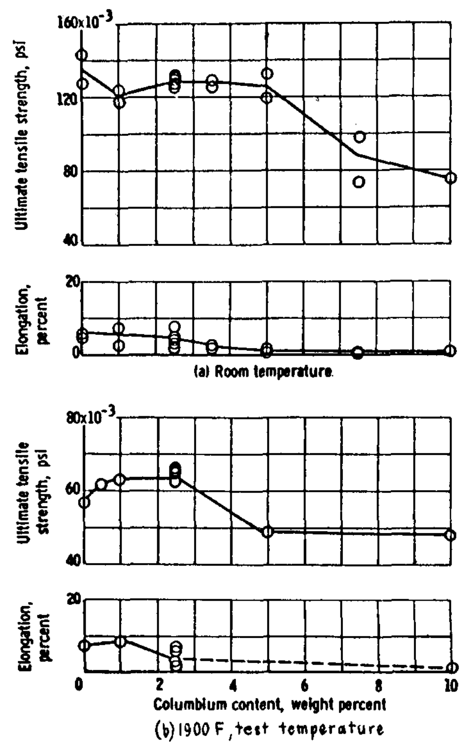


FIGURE 119.–Effect of columbium additions on as-cast tensile properties of modified TAZ-8A alloy.

However, the room-temperature strength of forged TAZ-8 is greater than as-cast. The cast material has better ductility at all test temperatures. In research on modifications of TAZ-8, the same trend was followed by TAZ-8 containing 1.80% V, instead of the normal 2.5 percent (ref. 91). As the isolated points on figure 117 (ref. 89) indicate, the as-cast 1.8% V alloy exhibited lower ductility than did the normal

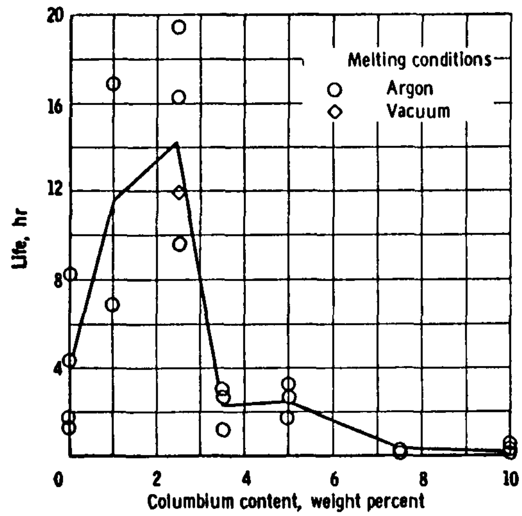


FIGURE 120.–Effect of columbium additions on stress-rupture life at 2000° F and 15,000 psi of TAZ 8A alloy.

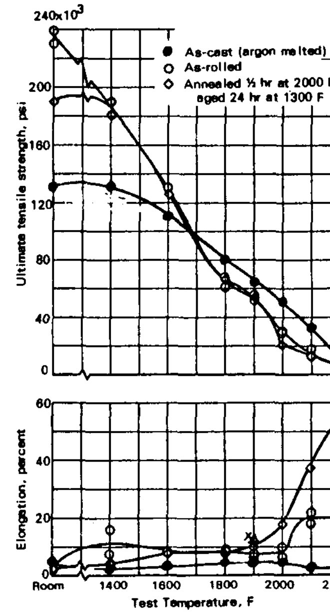


FIGURE 121.–Tensile properties of cast TAZ-8A alloy sheet showing the effect of heat treatment and hot rolling.

2.5% V alloy. The tensile strength is also slightly different at these two vanadium levels.

An experimental heat of TAZ-8, modified with lower vanadium (1.63%) and higher carbon (1.73%), exhibited rapid work hardening as illustrated by figure 118 (ref. 85). This, and two other heats, show the benefit of various heat treatments in table 99 (ref. 85). Although only a small amount of data was obtained, they suggest that a definite improvement in strength can be obtained as a result of aging at 1600° and 1450° F.

Walters and Freche (ref. 92) showed that by substituting columbium for vanadium the oxidation resistance could be improved without loss in strength or workability. The new alloy, designated TAZ-8A, consists of (percent):

Ni:	balance (68)	Mo:	4
Cr:	6	W:	4
Ta:	8	Zr:	1
Cb:	2½	B:	0.004
Al:	6	C:	0.125

The effect of columbium on the strength and ductility at room temperature and at 1900° F is illustrated in figure 119 (ref. 92). Strength at both temperatures was greatest at 2.5% columbium, but the room-temperature ductility, although still adequate, was reduced. Figure 120 (ref. 92) shows that the

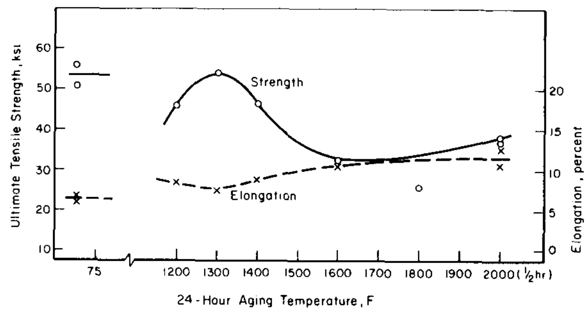


FIGURE 122.–Room-temperature tensile properties of hot-rolled TAZ-8 alloy at 1900° F showing the effect of aging temperature. (Material was annealed at 2000° F/½ hr, AC prior to aging.)

creep-rupture life (2000° F/15,000 psi) also seemed to be highest at the 2.5% columbium level.

The effect of hot rolling and of heat treating on the strength and ductility of TAZ-8A at ambient and at elevated test temperatures is shown in figure 121 (ref. 92). The as-cast material exhibited superior strength but inferior ductility above 1700° F, while the worked and heat-treated alloys showed much higher strength at 1600° F and below. Annealing at

2000° F and aging at 1300° F lowered the tensile strength of wrought material near ambient temperatures but greatly improved the ductility at 2000° F and above. Hence, the preferable condition of the material would depend on the service temperature and the expected mode of failure or design critiera. The cast condition would seem suitable for service between 1700° and 2100° F under static conditions of stress. The effect of aging temperature on the tensile strength and ductility of hot-rolled TAZ-8A at 1900° F is shown in figure 122 (ref. 92). No strength improvement over the as-rolled material was gained, even at the optimum of 1300° F, and the ductility improved only slightly, i.e., from 7-1/2 to 10 percent.

TABLE 4—*Mechanical Properties of Wrought Nickel in Various Forms and Conditions (ref. 1)*

Form and Temper	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 2 In., percent	Hardness, Brinell (3000 kg)	Hardness, Rockwell B
Wire					
Cold-drawn					
Annealed	55-85	15-50	50-30	–	–
No. 0 temper	80-95	–	–	–	–
No. 1 temper	70-95	40-75	40-20	–	–
Regular temper	105-140	105-135	15-4	–	–
Spring temper	125-145	105-135	15-2	–	–
Rod and bar					
Cold-drawn					
Annealed	55-80	15-30	55-40	90-120	45-70
As-drawn	65-110	40-100	35-10	140-230	75-98
As-rolled	60-85	15-45	55-35	90-150	45-80
Forged, hot-finished	65-90	20-60	55-40	100-170	55-85
Plate					
Hot-rolled					
Annealed	55-80	15-40	60-40	90-140	45-75
As-rolled	55-85	20-50	55-35	100-150	55-80
Sheet					
Annealed	55-75	15-30	55-40	–	70 max
Hard	90-115	70-105	15-2	–	90 min
Strip					
Annealed	55-75	15-30	55-40	–	64 max
Spring	90-130	70-115	15-2	–	95 min
Tubing					
Cold-drawn					
Annealed	55-75	20-30	60-40	–	70 max
Hard, stress relieved	65-110	40-90	35-15	–	75-98
Condenser and evaporator					
Annealed	55-75	15-30	60-40	–	65 max
Stress-relieved	65-90	40-75	35-20	–	75-92

TABLE 5—*Compression, Tension, and Hardness Data for Nickel Related to Worked Conditions (ref. 1)*

[All strength data taken to nearest 1000 psi]

	Hot-Rolled	Cold-Drawn, 24 Percent	Annealed
Compression data			
Yield strength (0.2% offset), 1000 psi	23	58	26
Proportional limit, 1000 psi	18	41	15
Tensile data			
Breaking strength, 1000 psi	71	87	73
Yield strength (0.2% offset), 1000 psi	24	62	27
Proportional limit, 1000 psi	18	52	19
Hardness data			
Brinell (3000 kg)	107	177	109

TABLE 6—*Shear Strength of Annealed and Hard Nickel (ref. 1)*

[All strength data taken to nearest 1000 psi]

Temper	Shear Strength, 1000 psi	Tensile Strength, 1000 psi	Hardness, Rockwell B
Annealed	52	68	46
Half-hard	58	79	90
Full-hard	75	121	100

TABLE 7—*Fatigue and Corrosion-Fatigue Limits of Nickel (ref. 1)*

No. of Cycles	Stress to Cause Failure[a], 1000 psi					
	Cold-Drawn Rod Tested in			Annealed Rod Tested in		
	Air	Fresh Water	Seawater	Air	Fresh Water	Seawater
10^4	109	110	–	–	–	–
10^5	84	80	–	52	52	52
10^6	63	56	54	40	39	37
10^7	52	34	30	34	27	24
10^8	50	26	23	33	23	21
10^9	50	24	21	33	23	21

(a) The stress given is the stress in one direction only; the method of testing causes stress to vary from these values in tension to the same values in compression, and return, during each cycle of operation

TABLE 8—*Temper and Hardness of Cold-Rolled Nickel 200 and Low-Carbon Nickel 201 Sheet and Strip (ref. 1)*

Condition or Temper	Hardness, Rockwell B	
	Nickel 200	Nickel 201
	Sheet	
Spinning quality	64 max	55 max
Deep drawing	64 max	55 max
Annealed	70 max	62 max
1/4-hard	70-80	–
1/2-hard	79-86	–
Hard	90 min	–
	Strip	
Spinning quality	64 max	55 max
Deep drawing	64 max	55 max
Annealed	64 max	55 max
Skin hard	64-70	–
1/4-hard	70-80	–
1/2-hard	79-86	–
3/4-hard	85-91	–
Hard	90-95	–
Spring (full hard)	95 min	–

TABLE 9—*Tensile Properties, Modulus of Elasticity, and Creep Strength of Annealed Nickel at Elevated Temperatures*[a]

Temperature, °F	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 2 In., percent	Modulus of Elasticity, 10^6 psi	Creep Strength, 1000 psi, Stress to Produce a Minimum Creep Rate of		Stress-Rupture Properties Stress, 1000 psi, to Produce Rupture in		
					0.01 Percent in 1000 hr	0.1 Percent in 1000 hr	1000 hr	10,000 hr	100,000 hr
				Standard Wrought Nickel (Nickel 200)[b]					
Room[c]	67	21	47	30	–	–	–	–	–
200[c]	66	22	46	29	–	–	–	–	–
300[c]	66	21	44	29	–	–	–	–	–
400[c]	66	20	44	28	–	–	–	–	–
500[c]	67	19	45	28	–	–	–	–	–
600[c]	66	20	47	27	13	40	–	–	–
650	62	18	47	–	7	22	–	–	–
700	52	17	61	27*	3	11	–	–	–
750	–	–	–	–	2	8	–	–	–
800	44	16	65	26*	2	6	–	–	–
900	37	15	66	24*	–	–	–	–	–
1000	31	13	69	22*	–	–	–	–	–
1100	26	11	72	21*	–	–	–	–	–
1200	21	10	76	18*	–	–	–	–	–
1400	14	7	89	15*	–	–	–	–	–

TABLE 9–*Tensile Properties, Modulus of Elasticity, and Creep Strength of Annealed Nickel at Elevated Temperatures*(a)–Concluded

Temperature, °F	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 2 In., percent	Modulus of Elasticity, 10^6 psi	Creep Strength, 1000 psi, Stress to Produce a Minimum Creep Rate of		Stress-Rupture Properties Stress, 1000 psi, to Produce Rupture in		
					0.01 Percent in 1000 hr	0.1 Percent in 1000 hr	1000 hr	10,000 hr	100,000 hr
Low-Carbon Nickel (Nickel 201)									
Room(c)	58	–	50	30	–	–	–	–	–
200(c)	56	20	45	29	–	–	–	–	–
300(c)	54	15	45	29	–	–	–	–	–
400(c)	54	18	43	28	–	–	–	–	–
500(c)	54	16	41	28	–	–	–	–	–
600(c)	52	18	42	27	–	–	–	–	–
700(c)	47	14	–	26	14	25	–	–	–
800(c)	41	13	–	26	7	13	27	–	–
900(c)	37	13	–	25	4	7	20	–	–
1000(c)	33	12	–	25	2	4	14	11	8
1100(c)	27	11	–	24	2	3	11	8	6
1200(c)	22	9	–	23	1	2	7	5	3

(a) Above data represents short-time, high-temperature tests made in accordance with ASTM Designations E8-46 and E21-43; modulus of elasticity values were determined by vibratory method of Roberts and Northcliffe; *Journal of Iron and Steel Institute* (November, 1945), p 345, except those asterisked values which are secured from stress-strain diagrams

(b) Not recommended for stress applications above 600° F; use low-carbon nickel for these applications

(c) ASME code limiting temperature

TABLE 10–*Some Mechanical Properties of Nickel at Low Temperatures (ref. 1)*

Condition	Temperature, °F	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation in 2 In., percent	RA,(2) percent	Hardness, Rockwell C
Hot-rolled	RT(1)	24	65	50.0	–	–
Hot-rolled	-112	27	76	–	–	–
Hot-rolled	-292	28	98	–	–	–
Hot-rolled	-310	–	103	51.0	–	–
Cold-drawn	RT	97	103	16.3	66.9	19
Cold-drawn	-110	101	112	21.5	60.9	22

(1)Room temperature
(2)Reduction in area

TABLE 11–*Impact Strength of Nickel in Various Conditions at Very Low Temperatures (ref. 24)*

Condition	Impact Energy, ft-lb			Specimen
	80° F	-110° F	-300° to -315° F	
Annealed	216	235	234	Charpy-V
Hot-rolled	195	236	227	Charpy-V
Cold-drawn	185	205	210	Charpy-V
Cold-drawn	204	216	–	Charpy-V

TABLE 12–*Nominal Mechanical Properties of Monel K-500 in Various Forms and Conditions (ref. 25)*

Form and Condition	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation, percent	Hardness	
				Brinell (3000 kg)	Rockwell
Rod and bar					
Hot-finished	90-155	40-110	45-20	140-315	75B-35C
Hot-finished, aged[a]	140-190	100-150	30-20	265-346	27-38C
Hot-finished, annealed	90-110	40-60	45-25	140-185	75-90B
Hot-finished, annealed and aged[a]	130-165	85-120	35-20	250-315	24-35C
Cold-drawn, as-drawn	100-140	70-125	35-13	175-260	88B-26C
Cold-drawn, aged[a]	135-185	95-160	30-15	255-370	25-41C
Cold-drawn, annealed	90-110	40-60	50-25	140-185	75-90B
Cold-drawn, annealed and aged[a]	130-190	85-120	30-20	250-315	24-35C
Sheet, cold-rolled, annealed	90-105	40-65	45-25	–	85B max
Strip, cold-rolled					
Annealed	90-105	40-65	45-25	–	85B max
Annealed and aged[a]	130-170	90-120	25-15	–	24C min
Spring temper	145-165	130-160	8-3	–	25C min
Spring temper, aged[a]	170-220	130-195	10-5	–	34C min
Tube and pipe, seamless					
Cold-drawn, annealed	90-110	40-65	45-25	–	90B max
Cold-drawn, annealed and aged[a]	130-180	85-120	30-15	–	24-36C
Cold-drawn, as-drawn	110-160	85-140	15-2	–	95B-32C
Cold-drawn, as-drawn, aged[a]	140-220	100-200	25-3	–	27-40C
Plate					
Hot-finished	90-135	40-110	45-20	140-260	75B-26C
Hot-finished, aged[a]	140-180	100-135	30-20	265-337	27-37C
Wire, cold-drawn[b]					
Annealed	80-110	35-65	40-20	–	–
Annealed and aged[a]	120-150	90-110	30-15	–	–
Spring temper	145-190	130-180	5-2	–	–
Spring temper, aged[a]	160-200	140-190	8-3	–	–

(a) Nominal properties for material age hardened to produce maximum properties
(b) Properties shown are for sizes 0.0625 to 0.250-in. diam; properties for other sizes may differ from these

TABLE 13–*Effect of Short-Time Aging Procedures on Tensile Properties of Monel K-500[a] (ref. 25)*

Condition	Thermal Treatment		Tensile Properties			Hardness, Rockwell C
	Temperature, °F	Time, hr	Yield Strength, (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	
Rod, hot-rolled	–	0	45	93	44	82B
	1100	2	82	132	36	17
	1100	4	86	136	34	20
	1100	8	90	142	33	22
Strip, annealed	–	0	50	100	39	85B
	1100	2	90	142	31	24
	1100	4	96	141	27	25
	1100	8	98	140	27	26

TABLE 13–*Effect of Short-Time Aging Procedures on Tensile Properties of Monel K-500*[a] *(ref. 25)*–Concluded

Condition	Thermal Treatment: Temperature, °F	Thermal Treatment: Time, hr	Tensile Properties: Yield Strength, (0.2% Offset), 1000 psi	Tensile Properties: Tensile Strength, 1000 psi	Tensile Properties: Elongation, percent	Hardness, Rockwell C
Strip, cold-rolled 10%	–	0	90	111	27	19
	1100	2	122	155	23	31
	1100	4	122	155	21	31
	1100	8	123	156	21	31
	1000	2	124	141	24	31
	1000	4	123	144	23	31
	1000	8	129	149	22	32
Strip, cold-rolled 20%	–	0	115	125	14	23
	1100	2	140	163	18	34
	1100	4	142	163	18	33
	1100	8	141	163	18	33
	1000	2	143	169	17	34
	1000	4	143	170	18	34
	1000	8	148	174	18	35
Strip, cold-rolled 40%	–	0	136	143	5	27
	1100	2	159	175	14	37
	1100	4	159	176	14	36
	1100	8	156	174	14	36
	1000	2	165	182	11	37
	1000	4	164	183	14	37
	1000	8	167	184	13	38
Strip, cold-rolled 50%	–	0	141	148	4	29
	1100	2	166	179	12	38
	1100	4	165	181	12	38
	1100	8	161	177	13	38
	1000	2	173	187	10	39
	1000	4	174	189	13	39
	1000	8	174	189	11	39

(a) These data are offered as a guide to short-time aging treatments and are not suitable for specification purposes

TABLE 14–*Shear Strength of Monel K-500*[a] *(ref. 25)*

Condition	Maximum Strength, 1000 psi	Deflection at Maximum Strength in./in.	Tensile Strength, 1000 psi	Elongation, percent	Hardness, Rockwell C
Annealed	65	0.08	97	49.0	84B
Annealed, aged	96	0.06	147	29.0	29
Half-hard	71	0.04	122	12.5	25
Half-hard, aged	98	0.05	155	24.0	31
Full-hard	89	0.04	151	16.5	33
Full-hard, aged	98	0.04	168	12.5	37

(a) The tests were made in double shear

TABLE 15–*Bearing Strength of Monel K-500*[(a)] *(ref. 25)*

Condition	Tensile Properties			Bearing Strength		Ratio, Bearing to	
	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation, percent	Ultimate Strength (Tearing Out), 1000 psi	Yield Strength (2 Percent Enlargement of Hole Diameter in Sheet), 1000 psi	Ultimate Strength	Yield Strength
Annealed	92	38	49.0	178	68	1.93	1.79
Annealed, aged	145	98	31.0	295	162	2.03	1.65
Hard	145	139	5.0	249	190	1.72	1.37
Hard, aged	195	177	10.0	358	262	1.83	1.48

(a) Bearing strength data were determined with 0.062 x 1.25 x 2.5-in. material having a 3/16-in. hole drilled 3/8 in. from the edge; the pin fitted closely into the hole; the maximum load for tearing out of the hole and the load required for a permanent enlargement of the hole diameter by 2% were determined and calculated as ultimate and yield strengths, respectively, in the bearing

TABLE 16–*Compressive Strength of Monel K-500 (ref. 25)*

Property	Hot-Rolled		Cold-Drawn	
	As-Rolled	Aged	As-Drawn	Aged
Hardness				
Brinell (3000 kg)	165	300	205	330
Rockwell C	5	33	23	35
Vickers (30 kg–diamond pyramid)	167	316	210	336
Tension				
Tensile strength, 1000 psi	100	151	106	158
Yield strength (0.2% offset), 1000 psi	47	111	85	120
Elongation, percent	42.5	30.0	26.5	22.0
Compression				
Yield strength (0.2% offset), 1000 psi	40	121	76	121
Yield strength (0.01% offset), 1000 psi	34	96	55	102

TABLE 17–*Torsional Properties of Monel K-500 (ref. 25)*

Condition	Yield Strength (0.00 Percent Offset)[(a)], 1000 psi	Johnson's Apparent Elastic Limit, 1000 psi	Angle of Twist, deg/in.
Hot-rolled	27	29	620
Hot-rolled, aged	57	67	104
Cold-drawn	48	55	360
Cold-drawn, aged	62	71	76

(a) $Ss = \frac{5.08\ Mt}{d^3}$,

where Ss = torsional stress on the outer fiber, psi
Mt = torsional moment, in.-lb
d = specimen diameter, in.

TABLE 18–*Room-Temperature Impact Data for Monel K-500 (ref. 25)*

Condition	Test Orientation	Charpy Keyhole Impact Strength, ft-lb
Hot-finished	Longitudinal	74
	Transverse	51
Hot-finished, annealed[(a)]	Longitudinal	75
	Transverse	48
Hot-finished, aged[(b)]	Longitudinal	39[(c)]
	Transverse	23[(c)]
Hot-finished, aged[(c)]	Longitudinal	25[(c)]
	Transverse	20[(c)]
Hot-finished, annealed and aged[(e)]	Longitudinal	38[(c)]
	Transverse	22[(c)]
Cold-drawn	Longitudinal	40
Cold-drawn, annealed[(a)]	Longitudinal	90
Cold-drawn, aged[(b)]	Longitudinal	26[(c)]
Cold-drawn, aged[(d)]	Longitudinal	20[(c)]
Cold-drawn, annealed and aged[(e)]	Longitudinal	46[(c)]

(a) 1800° F/1 hr, water quench
(b) 1100° F/16 hr, air-cool
(c) Specimen fractured completely
(d) 1100° F/16 hr, *FC 15°/hr to 900°F
(e) Anneal (a) plus age (d)

*FC–furnace cool

TABLE 19–*Data on the Fatigue Resistance of Monel K-500*

Condition	Endurance Limit (10^8 Cycles), 1000 psi	Tensile Strength, 1000 psi	Endurance Ratio
	Endurance Limits of Rod[(a)]		
Annealed	38	88	0.43
Hot-rolled	43	99	0.43
Hot-rolled, aged	51	155	0.33
Cold-drawn	45	120	0.37
Cold-drawn, aged	47	170	0.28
	Endurance Limit of Strip[(b)]		
Annealed	27	88	0.31
Spring temper, aged	37	153	0.24

(a) The data for rod were developed on high-speed rotating-beam fatigue machines using smooth polished specimens.
(b) Specimens were subjected to alternate back and forth bending as a flat spring; the samples were cut with the longitudinal direction parallel to the direction of rolling.

TABLE 20–*Short-Time Elevated-Temperature Tensile Properties of Monel K-500 Rod in Several Conditions (ref. 25)*

Temperature °F	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation percent
Rod, Hot-Rolled, As-Rolled			
70	97	49	44.0
200	96	45	42.5
400	92	44	42.0
600	94	36	39.5
800	87	39	29.5
1000	72	42	13.5
1200	63	44	2.5
1400	36	17	22.5
1600	15	–	53.0
1800	10	–	65.5
Rod, Hot-Rolled, Age-Hardened			
70	160	111	23.5
200	150	108	23.5
400	149	103	24.0
600	146	105	23.0
800	124	105	8.5
1000	95	92	3.0
1200	80	80	1.5
1400	45	30	8.0
1600	21	–	23.0
1800	6	–	47.5
2000	3	–	81.5
Rod, Cold-Drawn, Annealed, and Age-Hardened			
70	156	105	23.0
200	152	101	22.0
300	148	98	22.0
350	147	96	21.0
400	147	96	22.0
450	144	98	21.0
500	143	94	21.0
550	143	96	20.0
600	142	96	20.0
700	138	95	20.0

TABLE 21–*Stress-Rupture Strength of Monel K-500 (ref. 25)*

Condition	Temperature, °F	Stress, 1000 psi, to Produce Rupture in		
		100 hr	1000 hr	10,000 hr
Hot-finished, aged	900	66	46	33
	1000	42	30	21
	1100	34	27	20

TABLE 22–*Creep Strength of Monel K-500 (ref. 25)*

Condition	Temperature, °F	Stress, 1000 psi, to Produce a Creep Rate of 0.10 percent in 10,000 hr	Stress, 1000 psi, to Produce a Creep Rate of 1.00 percent in 10,000 hr
Cold-drawn, aged	750	68	–
	800	47	88
	900	25	48
	1000	8	21
	1100	–	9

TABLE 23–*Endurance Limit of Monel K-500 at 1000° F (ref. 25)*

Condition	Temperature °F	Endurance Limit (10^8 Cycles), 1000 psi
Hot-finished, aged	80	46
	1000	43
Cold-drawn, aged	80	52
	1000	48

TABLE 24–*Hot Hardness of Monel K-500 (ref. 25)*

Condition	Hardness, Brinell, at Temperatures, °F: 70	700	800	900	1000	1100
Hot-finished	241	223	207	201	170	179
Hot-finished, aged	331	311	302	293	255	229

TABLE 25–*Tensile Properties of Monel K-500 Sheet at Cryogenic Temperatures (ref. 26)*

[0.020-in. sheet, age hardened 1080° F/16 hr]

Test Temperature °F	Direction	Yield Strength, 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Notched-Unnotched Ratio, $K_t = 6.3$
78	Long.	97.3	154	22	0.93
-100	Long.	107	166	24	0.93
-320	Long.	120	183	30	0.95
-423	Long.	136	200	28	0.99

TABLE 26–*Bending Impact Strength of Monel K-500*

Condition	Average Energy Absorbed, ft-lb: Smooth Specimen			Notched Specimen, $K_t = 4.00$		
	-200° F	-120° F	+80° F	-200° F	-120° F	+80° F
Hot-finished, aged	–	–	–	42	50	55
Cold-drawn, aged	–	–	185	30	30	32

TABLE 27–*Tension Impact Strength of Monel K-500*

Condition	Average Energy Absorbed, ft-lb					
	Smooth Specimen			Notched Specimen, $K_t = 3.00$		
	-200° F	-120° F	+80° F	-200° F	-120° F	+80° F
Hot-finished, aged	158	145	141	37	37	35
Cold-drawn, aged	127	108	117	34	28	29

TABLE 28–*Fatigue Strength of Low Temperatures of Monel K-500*

Temperature, °F	Stress, 1000 psi, for a Fatigue Life of		
	10^5 Cycles	10^6 Cycles	10^7 Cycles
70	90	55	37
-110	99	67	–
-320	105	69	–
-423	143	101	–

TABLE 29–*Room and Hot Tensile Properties of Mill-Annealed (1750° F) Inconel X-750 (ref. 28)*

Temperature °F	Tensile Strength, 1000 psi	Yield Strength, 1000 psi, Offset		Elongation in 2 In., percent	Reduction in Area, percent
		0.02%	0.2%		
		1 In. Round			
85	174	108	115	27.5	41.0
400	168	110	113	29.0	40.0
600	160	101	107	30.0	37.0
800	159	105	106	31.0	42.0
		7/8 In. Round			
85	168	108	114	65.0	49.5
400	154	104	104	35.0	40.0
600	149	96	98	36.0	39.5
800	147	95	99	33.0	43.5

TABLE 30–*Age-Hardening Procedures for Inconel X-750*

Treatment	Solution Treating, Stress Equalizing, or Annealing(a)	Aging Conditions(b)	Total Heating Time, hr	Remarks
1	2100° F/2 to 4 hr, AC	1550° F/24 hr, AC plus 1300° F/20 hr, AC	46-48	Service above 1100° F, hot-rolled bar
2	1625° F/24 hr, AC	1300° F/20 hr, AC	44	–
3	1750° F/1 hr, AC	1350° F/8 hr, FC 25° F/hr to 1150° F, AC	17	Service below 1100° F
4	1750° F/1 hr, AC	1350° F/8 hr, FC at up to 200° F/hr to 1150° F, hold to a total aging time of 16 hr	17	Service below 1100° F

TABLE 30–*Age-Hardening Procedures for Inconel X-750*–Concluded

Treatment	Solution Treating, Stress Equalizing, or Annealing[a]	Aging Conditions[b]	Total Heating Time, hr	Remarks
5	1750° F/1 hr, AC	1400° F/1 hr, FC 50° F/hr to 1150° F, AC	7	Slightly lower properties than treatment 3
6	–	1400° F/1 hr, FC 50° F/hr to 1150° F, AC	6	–
7	–	1350° F/8 hr, FC 25° F/hr to 1150° F, AC	16	–
8	–	1300° F/20 hr, AC	20	–

(a) AC–air cool
(b) FC–furnace cool

TABLE 31–*Short-Time Aging Treatment for Inconel X-750 Rod Followed by Air Cooling (ref. 18)*

Condition	Aging		Tensile Properties			Hardness, Rockwell C
	Temperature °F	Time, hr	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation in 2 In. percent	
	1300	0	77	143	36	23
Hot-rolled	1300	1	110	169	29	33
and	1300	2	114	173	28	34
equalized	1300	5	120	179	26	36
1625° F/4 hr	1300	10	125	181	25	36
	1300	20	125	182	23	37

TABLE 32–*Short-Time Aging Treatment for Annealed 0.065-In. Inconel X-750 Sheet Followed by Air Cooling (ref. 18)*

Condition	Aging		Tensile Properties				Hardness, Rockwell C
	Temperature °F	Time, hr	Proof Stress (0.01% Offset), 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength 1000 psi	Elongation in 2 In. percent	
	0	0	31	45	111	47	8
	1300	1	85	95	155	35	27
	1300	2	90	107	165	28	31
	1300	4	96	115	172	26	33
	1300	8	100	120	173	24	34
	1300	20	106	122	176	23	35
	1350	1	92	104	163	29	30
Cold-rolled	1350	2	97	111	169	26	32
and	1350	4	98	116	172	25	33
annealed	1350	8	102	120	174	23	34
20 min at							
2000° F	1400	1	96	109	167	29	32
	1400	2	98	113	170	24	33
	1400	4	99	115	171	23	33
	1450	1	90	106	165	29	31
	1450	2	98	111	166	28	31

TABLE 32–*Short-Time Aging Treatment for Annealed 0.065-In. Inconel X-750 Sheet Followed by Air Cooling (ref. 18)*–Concluded

Condition	Aging		Tensile Properties				Hardness, Rockwell C
	Temperature °F	Time, hr	Proof Stress (0.01% Offset), 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength 1000 psi	Elongation in 2 In. percent	
	1500	1	87	104	161	27	29
	1500	2	92	105	166	29	31
	1600	1	71	83	140	32	22

TABLE 33–*Room-Temperature Tensile Properties of Hot-Rolled Bars of Inconel X-750 (ref. 28)*

Heat Treatment[a]	Diam, in.	Yield Strength, 1000 psi, Offset		Tensile Strength, 1000 psi	Elongation, percent	Reduction in Area, percent	Hardness, Rockwell C
		0.02%	0.2%				
5	1/2	136	146	199	25.0	41.5	36.0
3		144	149	196	24.0	42.3	38.0
5	21/32	128	139	194	27.0	46.4	38.0
3		128	139	192	25.0	47.7	39.0
5	3/4	127	137	193	25.0	38.5	38.0
3		131	140	191	22.0	38.8	39.0
5	7/8	129	140	194	24.0	40.2	40.0
3		137	146	197	21.0	42.8	40.0
5	1	126	130	187	25.0	41.8	33.0
3		132	139	190	22.0	35.4	39.4
5	1-3/16	126	134	189	24.0	39.5	39.0
3		129	137	192	23.0	41.0	40.0
5	1-3/16	127	132	195	25.0	43.2	35.0
3		133	138	195	26.0	43.5	42.0
5	1-3/8	130	136	190	24.0	43.0	38.0
3		130	136	190	23.0	43.0	37.0
5	1-1/2	130	132	188	27.0	46.0	34.0
3		130	132	189	26.0	45.0	40.0
5	1-1/2	138	141	198	24.0	42.0	41.0
3		138	142	196	25.0	46.3	40.0
5	1-1/2	123	129	190	26.0	43.0	40.0
3		123	131	190	25.0	40.5	41.0
5	2-1/4	127	138	189	22.0	30.5	39.0
3		128	140	189	21.0	21.5	39.0

TABLE 33–*Room-Temperature Tensile Properties of Hot-Rolled Bars of Inconel X-750 (ref. 28)*–Concluded

Heat Treatment(a)	Diam in.	Yield Strength, 1000 psi, Offset		Tensile Strength, 1000 psi	Elongation, percent	Reduction in Area, percent	Hardness, Rockwell C
		0.02%	0.2%				
5	2-1/2	124	135	184	23.0	38.0	38.0
3		126	137	184	22.0	36.0	39.0
5	2-15/16	117	128	180	24.0	35.0	34.0
3		122	137	184	23.0	38.0	38.0

(a) See table 30 for details of heat treatment

TABLE 34–*Room-Temperature Tensile Data for Cold-Rolled Annealed Inconel X-750 Sheet After Aging by Two Different Treatments (ref. 28)*

Heat Treatment(a)	Thickness, in.	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Hardness, Rockwell C
8	0.012	113	175	25.0	35.0(b)
7		126	183	22.0	35.0(b)
8	0.015	126	182	24.0	37.0(b)
7		129	189	23.0	39.0(b)
8	0.025	116	178	28.0	37.0(b)
7		127	186	22.0	39.0(b)
8	0.025	119	181	25.0	39.0(b)
7		129	188	23.0	39.0(b)
8	0.025	113	174	28.0	35.0(b)
7		124	182	24.0	37.0(b)
8	0.025	123	178	25.0	37.0(b)
7		131	184	24.0	41.0(b)
8	0.032	120	177	27.0	37.0(b)
7		125	183	26.0	39.0(b)
8	0.032	119	178	26.0	37.0(b)
7		133	187	23.0	39.0(b)
8	0.050	120	175	27.0	36.0(b)
7		130	182	25.0	39.0
8	0.055	131	184	26.0	39.0
7		138	189	23.0	40.0
8	0.062	125	180	26.0	36.0
7		131	185	26.0	38.0
8	0.062	124	181	26.0	38.0
7		133	189	23.0	41.0
8	0.062	129	186	26.0	39.0
7		139	193	23.0	41.0
8	0.062	130	181	26.0	39.0
7		137	189	23.0	40.0
8	0.062	122	178	27.0	37.0
7		134	187	24.0	39.0
8	0.068	124	180	27.0	38.0
7		135	189	24.0	41.0
8	0.068	122	180	28.0	38.0

TABLE 34–*Room-Temperature Tensile Data for Cold-Rolled Annealed Inconel X-750 Sheet After Aging by Two Different Treatments (ref. 28)*–Concluded

Heat Treatment[a]	Thickness, in.	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Hardness, Rockwell C
7		135	188	24.0	41.0
8	0.071	122	178	27.0	37.0
7		134	187	24.0	42.0
8	0.071	126	181	26.0	37.0
7		134	185	23.0	40.0
8	0.071	125	178	27.0	37.0
7		133	185	25.0	40.0
8	0.071	123	181	27.0	37.0
7		136	189	24.0	40.0
8	0.071	120	177	28.0	37.0
7		130	185	25.0	41.0
8	0.125	126	185	26.0	39.0
7		138	190	24.0	41.0
8	0.125	128	179	26.0	38.0
7		134	188	25.0	40.0
8	0.140	126	175	26.0	39.0
7		133	182	25.0	39.0

(a) Details in table 30
(b) 15N and 30N hardness values converted to RC hardness values

TABLE 35–*Effect of Cold Work on Tensile Properties of Inconel X-750 (ref. 30)*

Strain, percent	Ultimate Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation, percent
0	178	123	26.0
0	179	121	24.0
0	179	121	26.5
5	181	115	20.0
5	181	121	22.5
5	185	130	23.0
15	190	144	20.0
15	195	147	18.5
15	194	148	19.0
20	195	141	17.0
20	193	153	16.5
20	200	159	16.5
25	205	164	14.0
25	204	163	15.0
25	204	166	15.0

Note: Specimens strained in the mill-annealed condition and then aged at 1300° F for 20 hr and not descaled; 0.063 sheet, strained and tested in the longitudinal direction

TABLE 36–*Effect of Cold Work and Heat Treatment on Mechanical Properties of Inconel X-750 Wire (ref. 28)*

Condition and Thermal Treatment[(a)]	Tensile Strength, 1000 psi, Diam, in.		0.2% Offset Yield Strength, 1000 psi Diam, in.		Proportional Limit (0.0% Offset), 1000 psi, Diam, in.		Elongation in 2 In., percent Diam, in.		Modulus of Elasticity in Tension, psi x 10^6, Diam, in.		Modulus of Rigidity in Torsion[(a)] psi x 10^6, Diam, in.	
	0.020	0.229	0.020	0.229	0.020	0.229	0.020	0.229	0.020	0.229	0.020	0.229
No. 1 Temper Wire (15%)												
As-drawn	139	145	68	119	40	56	30.0	24	30.6	–	11.3	–
Annealed 1950° F/15 min, WQ	120	110	43	42	34	24	34.0	53	31.0	–	11.3	10.7
Solution-treated 2100° F/2 hr, WQ plus 1550° F/24 hr plus 1300° F/20 hr, and AC	158	166	93	101	65	64	13.0	14	30.2	–	11.7	11.1
Aged 1350° F/16 hr, AC	202	204	141	159	93	91	16.0	16	31.2	–	11.6	11.4
Aged 1200° F/4 hr, AC	178	176	109	136	77	81	25.0	19	30.4	–	11.7	11.4
Spring Temper Wire (65%)												
As-drawn	269	–	233	–	137	–	1.6	–	26.0	–	10.2	–
Annealed 1950° F/15 min, WQ	130	–	51	–	37	–	33.0	–	–	–	11.0	–
Solution-treated 2100° F/2 hr, WQ plus 1550° F/24 hr plus 1300° F/20 hr, and AC	154	–	104	–	81	–	8.0	–	30.8	–	11.2	–
Aged 1300° F/16 hr, AC	274	–	268	–	168	–	1.0	–	31.6	–	12.2	–
Aged 1200° F/4 hr, AC	298	–	293	–	173	–	1.0	–	31.0	–	11.9	–

(a) WQ–water quenched

TABLE 37–*Effect of Age-Hardening Conditions on Hot Hardness of Hot-Rolled Inconel X-750 (ref. 28)*

Treatment 1		Treatment 8[(a)]	
Test Temperature, °F	Hardness BHN	Test Temperature, °F	Hardness BHN
90	286	80	351
450	277	410	332
620	281	613	321
800	269	805	321
888	269	920	315
980	269	1008	311
1050	262	1095	304
1155	269	1200	309
1250	269	1300	311
1300	269	1390	298
1350	262	1460	269
1375	255	1530	255
1388	244	1560	217
1415	239	1605	179
1445	241	1705	131
1480	217		
1500	212		
1600	163		
1650	143		
1750	80		

(a) See table 30 for details

TABLE 38—*High-Temperature Tensile Properties for Cold-Rolled, Annealed Inconel X-750 Sheet After Aging by Two Different Treatments (ref. 28)*

Heat Treatment[a]	Test Temperature, °F	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Hardness, Rockwell C[b]
8	RT	122	177	27.0	37.0
7	RT	132	186	25.0	38.0
8	400	112	167	30.0	37.0
7	400	123	176	25.0	39.0
8	800	107	151	33.0	37.0
7	800	120	162	29.5	40.0
8	1000	112	154	26.0	38.0
7	1000	116	155	25.0	39.0
8	1100	105	135	10.5	37.0
7	1100	116	145	9.0	38.0
8	1200	105	123	6.0	38.0
7	1200	113	132	4.2	39.0
8	1300	100	110	3.5	37.0
7	1300	103	115	3.0	39.0
8	1500	76	80	11.0	37.0
7	1500	77	82	12.0	40.0

(a) Details in table 30
(b) Superficial hardness converted to R_C

TABLE 39—*High-Temperature Tensile Properties for 3/4-In. Diam Hot-Rolled Round Inconel X-750 (ref. 28)*

Heat Treatment[a]	Test Temperature, °F	Yield Strength (0.2 Percent Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Reduction in Area, percent	Hardness, Rockwell C
2	RT	126	184	25.0	41.5	39.0
3		140	195	24.0	40.5	37.0
2	600	116	169	23.0	35.0	37.0
3		131	178	21.0	41.0	37.0
2	800	114	166	24.0	39.0	37.0
3		131	173	21.0	38.0	37.0
2	1000	115	163	20.0	25.0	37.0
3		128	168	13.0	18.0	37.0
2	1100	112	159	10.0	13.0	37.0
3		126	157	8.0	11.0	39.0
2	1200	110	143	7.0	7.8	35.0
3		122	143	6.0	8.0	36.0
2	1350	98	107	6.0	10.0	38.0
3		107	114	5.0	8.0	36.0
2	1500	64	65	17.0	19.5	37.0
3		76	77	10.0	13.5	37.0

(a) See table 30 for details

TABLE 40–*Effect of Prior Exposure on the Tensile Properties of Cold-Rolled 0.025-In. Inconel X-750 Sheet at 650° F*

Prior Exposure[a]			Test Temperature, °F	Unnotched (Smooth) Specimens						Sharp-Edge Notches			
Temperature, °F	Stress, 1000 psi	Time, hr		Tensile Strength, 1000 psi		Yield Strength, (0.2% Offset), 1000 psi		Elongation in 2 In., percent		Tensile Strength, 1000 psi		Strength Ratio N/S[c]	
				L[b]	T	L	T	L	T	L	T	L	T
Cold-Rolled 27 Percent													
None			650	136.0		118.6		15.5		131.6		0.97	
650	40	1000	650	128.6		112.1		18		113.9		0.88	
Cold-Rolled 67 Percent													
None			650	194.8	195.2	156.0	156.0	1	3.5	153.8	139.1	0.79	0.71
650	40	1000	650	197.1	198.1	195.4	117.4	2.5	4.5	162.6	149.8	0.82	0.76

(a) Conditions of exposure prior to testing
(b) L–longitudinal orientation
T–transverse orientation

(c) $N/S = \dfrac{\text{Tensile strength of notched specimen}}{\text{Tensile strength of unnotched (smooth) specimen}}$

TABLE 41–*Effect of Heat Treatment and Test Temperature on Impact Strength (Charpy V-Notch Test, Ft-Lb) of Hot-Rolled Bars (ref. 28)*

Test Temperature, °F	Heat Treatment: 2100° F/2 hr, 1550° F/24 hr, 1300° F/20 hr[a]	1300° F/20 hr[b]
-320	33	34
-109	36	37
75	37	38
400	42	44
800	50	46
1000	49	49
1200	45	43
1350	49	49
1500	67	53
1600	113	82

(a) Treatment 1, table 30
(b) Treatment 8, table 30

TABLE 42–*Tensile Properties of Inconel X-750 at Cryogenic Temperatures (ref. 26)*
[0.063-in. sheet, annealed, aged 1300° F/20 hr, AC]

Test Temperature, °F	Direction	Yield Strength, 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Notched-Unnotched Ratio, $K_t = 6.3$
78	Longitudinal	118	174	25	0.97
	Transverse	118	174	25	0.97

TABLE 42—*Tensile Properties of Inconel X-750 at Cryogenic Temperatures (ref. 26)*—Concluded

[0.063-in. sheet, annealed, aged 1300° F/20 hr, AC]

Test Temperature, °F	Direction	Yield Strength, 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Notched-Unnotched Ratio, $K_t = 6.3$
-100	Longitudinal	122	189	30	0.92
-320	Longitudinal	130	214	31	0.86
	Transverse	130	212	30	0.87
-423	Longitudinal	134	233	30	0.85
	Transverse	139	234	31	0.86

TABLE 43—*Room- and Low-Temperature Tensile Properties of Inconel X-750 (ref. 28)*

[2100° F/2 hr, AC, 1550° F/24 hr, AC, 1300° F/20 hr, AC[a]]

Hardness, Rockwell C[b]	Temperature, °F	Tensile Strength, 1000 psi	Yield Strength[c], 1000 psi	Elongation in 1 In., percent	Reduction in Area, percent
		Smooth Specimens			
33	78	174	102	25.0	29.0
32	-104	186	115	23.0	26.0
34	-320	209	118	19.0	19.0
34	-423	208	130	15.0	15.0
		Notched Specimens[d]			
33	78	201	–	–	–
35	-104	200	–	–	–
35	-320	219	–	–	–
36	-423	225	–	–	–

(a) "Materials for Use at Liquid Hydrogen Temperature," ASTM Special Publication No. 287, 108 (1960)
(b) All hardness checks made at room temperature
(c) 0.2% offset except initial yield point at -423° F
(d) Notched bar (60-deg V notch, 0.037 in. deep, 0.005-in. radius at base)

TABLE 44—*Annealing and Aging Temperatures for Alloy 718 (ref. 13)*

Specification	Company	Annealing Temperature, °F	First Aging Temperature, °F	Second Aging Temperature, °F	Aging Method[a]
AMS 5596A	Society of Automotive Engineers	1750	1325	1150	I or II
B50T69-S6	General Electric Co.	1700	1325	1150	I
C50T79(S1)	General Electric Co.	1800	1325	1150	I
PWA 1009-C	Pratt & Whitney Aircraft Div.*	1750	1325	1150	I or II
EMS-581c	Airesearch Aviation Co.	1950	1350[b]	1200	I

TABLE 44–*Annealing and Aging Temperatures for Alloy 718 (ref. 13)*–Concluded

Specification	Company	Annealing Temperature, °F	First Aging Temperature, °F	Second Aging Temperature, °F	Aging Method(a)
RB0170-101	Rocketdyne Div.†	1950	1400	1200	III
AGC-44152	Aerojet-General Corp.	1950	1350	1200	IV

(a) I –Hold 8 hr at first aging temperature, furnace cool at 100° F/hr to second aging temperature; hold 8 hr, air cool
II –Hold 8 hr at first aging temperature, furnace cool to second aging temperature; hold at second aging temperature until total time elapsed since the beginning of the first aging is 18 hr
III–Hold 10 hr at first aging temperature, furnace cool to second aging temperature; hold at second aging temperature until total time elapsed since the beginning of the first aging is 20 hr
IV–Same as III, but first aging time may be 8 to 10 hr
(b) 1400° F on certain heavy forgings

*United Aircraft Corp.
†North American Rockwell Corp.

TABLE 45–*Effect of Solution-Treatment Temperature on Room-Temperature Tensile Properties of Allvac 718 Forged Bar (ref. 42)*

Treatment	Direction	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Yield Strength (0.02% Offset), 1000 psi	Elongation, percent	Reduction in Area, percent
1750° F solution plus double age(a)	L(b)	204	162	140	20.1	31.3
	T(c)	190	150	133	13.9	18.0
1800° F solution plus double age	L	200	159	138	21.4	32.4
	T	186	149	132	13.8	17.4
1950° F solution plus double age	T	177	145	129	19.7	23.7

(a) 1325° F/8 hr, FC to 1150° F, hold at 1150° F until total aging time equals 18 hr
(b) Longitudinal
(c) Transverse

TABLE 46–*Annealing of Overaged Alloy 718–Effect on Hardness and Grain Size (ref. 16)*
[Tests performed on 0.040-in. sheet]

Condition	Hardness, DPH	Average Grain Diameter, mm
Mill-annealed	188	–
30 hr/1400° F	358	–
30 hr/1400° F plus 15 min 1500° F, AC	318	0.025 - 0.035
30 hr/1400° F plus 15 min 1600° F, AC	215	0.025 - 0.035
30 hr/1400° F plus 15 min 1700° F, AC	188	0.025 - 0.035
30 hr/1400° F plus 15 min 1750° F, AC	186	0.025 - 0.035
30 hr/1400° F plus 15 min 1800° F, AC	185	0.025 - 0.035
30 hr/1400° F plus 15 min 1900° F, AC	179	0.025 - 0.035
30 hr/1400° F plus 15 min 2000° F, AC	157	0.120
30 hr/1400° F plus 15 min 2100° F, AC	156	0.150
30 hr/1400° F plus 15 min 2150° F, AC	151	0.150 - 0.200

TABLE 46—*Annealing of Overaged Alloy 718—Effect on Hardness and Grain Size (ref. 16)*—Concluded

[Tests performed on 0.040-in. sheet]

Condition	Hardness, DPH	Average Grain Diameter, mm
Mill annealed plus aged	438	–
30 hr/1400° F plus aged	401	–
30 hr/1400° F plus 15 min 1500° F plus aged	406	–
30 hr/1400° F plus 15 min 1600° F plus aged	424	–
30 hr/1400° F plus 15 min 1700° F plus aged	431	–
30 hr/1400° F plus 15 min 1750° F plus aged	437	–
30 hr/1400° F plus 15 min 1800° F plus aged	439	–
30 hr/1400° F plus 15 min 1900° F plus aged	442	–
30 hr/1400° F plus 15 min 2000° F plus aged	433	–
30 hr/1400° F plus 15 min 2150° F plus aged	427	–

TABLE 47—*Composition of Double-Vacuum-Cast Alloy 718 Hot-Rolled to 1/2-In. Bar, Fixed Composition: 18Cr, 18Fe, 3Mo, 0.4Co, 0.06C, Balance Ni (ref. 38)*

Heat Number	Content of Variable Elements, weight percent			
	Ti	Al	Cb	Si
39	0.91	0.68	5.41	0.17
40	0.94	0.67	5.41	1.03
41	0.37	0.74	5.34	0.60
42	0.97	0.75	5.47	0.61
43	1.37	1.12	6.47	0.92
44	0.52	0.31	6.03	0.95
45	1.20	0.34	4.03	0.89
46	0.56	1.22	3.94	0.88
47	1.26	0.28	6.00	0.30
50	0.68	1.05	5.93	0.34
51	1.45	1.28	3.85	0.34
52	0.52	0.26	3.94	0.31
53	0.90	0.64	6.53	0.63
54	0.96	0.61	3.64	0.64
55	1.45	0.64	5.31	0.61
56	0.90	1.36	5.25	0.65
57	0.94	0.06	5.17	0.60

TABLE 48–*Room-Temperature Tensile Properties of Wrought Alloy 718 as Affected by Cold Hydrostatic Extrusion and Aging (ref. 44)*

	Treatment Condition	Yield Strength (0.2% Offset), ksi	Ultimate Tensile Strength ksi	Elongation, percent	Reduction in Area, percent	Hardness, RC
A.	1800° F/4 hr, AC + 1325° F/8 hr, FC to 1150° F/8 hr, AC	153.4	198.5	20	30	40
B.	Treatment A + 50% cold hydrostatic extrusion	261.0	282.0	2	7	49
C.	Treatment B + 1150° F/8 hr, AC	282.0	287.0	3	10	51
D.	Treatment C + 15% cold hydrostatic extrusion + 1150° F/8 hr, AC	219.0	229.0	4	16	46
E.	1800° F/4 hr, AC + 1325° F/8 hr, FC + 50% cold extrusion + 1150° F/8 hr, AC	300.0	304.8	2	5	53
F.	1800° F/4 hr, + 50% cold extrusion + 1325° F/8 hr, FC to 1150° F/8 hr, AC	235.0	240.0	10	26	–
G.	1800° F/4 hr, + 50% cold-rolled + 1325° F/8 hr, FC to 1150° F/8 hr, AC	232.0	244.5	4	–	–

TABLE 49–*Compression Properties of Cast Alloy 718*

Condition	Yield Strength (0.2% Offset) ksi	Ultimate* Strength, ksi	Modulus of Elasticity psi
1800° F/2 hr, AC	25.0	–	6.3×10^6
1800° F/2 hr, AC	41.3	380	5.8×10^6
1800° F/2 hr, AC + 1325° F/8 hr, FC to 1150° F/8 hr, AC	84.0	366	6.4×10^6
1800° F/2 hr, AC + 1325° F/8 hr, FC to 1150° F/8 hr, AC	87.1	321	6.8×10^6

*Ductile shear failures; original area used to calculate stress

TABLE 50–*Hardness of Alloy 718 Wire, Cold-Drawn 64 Percent, with Various Aging Treatment*

Aging Treatment	Hardness Rockwell C
1450° F/8 hr, FC to 1300° F/total 20 hr, AC	45
1350° F/8 hr, FC to 1200° F/total 20 hr, AC	48
1250° F/8 hr, FC to 1150° F/total 20 hr, AC	50.6
1200° F/8 hr, FC to 1100° F/total 20 hr, AC	52.0
1125° F/8 hr, FC to 1000° F/total 20 hr, AC	52.1

TABLE 51–*Aging Treatments Used by Solar* for Alloy 718*

Total Aging Time, hr	Type	Aging Treatment
0.5	Single	1325° F/0.5 hr, AC
3.0	Single	1325° F/3 hr, AC
8.0	Single	1325° F/8 hr, AC
	Two-step	1325° F/4 hr, FC to 1150° F/4 hr, AC
16	Single	1325° F/16 hr, AC
18	Two-step	1325° F/8 hr, FC to 1150° F/10 hr, AC
20	Two-step	1400° F/10 hr, FC to 1200° F/10 hr, AC

*Division of International Harvestor Co.

TABLE 52–*Tensile Properties of Different Inconel 718 Products with Various Age-Hardening Schedules*

Test Temperature	Form and Size	Heat Treatment[(a)]	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Reduction in Area, percent	Hardness, Rockwell C
RT	Forged pancake	A	160	196	24.0	33.0	–
		S	147	188	24.0	34.0	–
	Hot-rolled, 5/8-in. diam	A	171	201	26.0	50.0	41
		B	160	193	20.0	52.0	42
		S	140	185	32.0	54.0	–
	Hot-rolled, 1/2-in. diam	A	174	211	23.0	40.0	–
		B	173	204	19.0	29.0	–
		S	154	193	23.0	36.5	–
	Cold-rolled annealed sheet 0.060-in. thick	C	155	185	13.0	–	41
		D	151	187	22.5	–	43
		T	137	176	23.5	–	–
	Forged rod 6-in. diam	B	152	184	28.0	42.0	–
		S	139	176	32.0	42.0	–
1200° F	Forged pancake	A	138	162	23.0	38.0	–
	Hot-rolled, 5/8-in. diam	A	145	164	28.0	59.0	–
	Hot-rolled, 1/2-in. diam	A	148	168	22.0	31.5	–
		S	132	153	12.0	12.0	–
	Cold-rolled annealed sheet 0.060-in. thick	C	126	149	13.0	–	–
		T	115	140	16.0	–	–
1300° F	Forged pancake	A	135	146	30.0	62.0	–
	Hot-rolled, 5/8-in. diam	A	133	145	22.0	34.0	–
	Hot-rolled, 1/2-in. diam	A	136	145	20.0	26.5	–
	Cold-rolled annealed sheet 0.060-in. thick	C	112	127	9.0	–	–
		T	109	121	6.0	–	–
	Forged rod 6-in. diam	B	113	129	14.0	16.0	–
1400° F	Forged pancake	A	109	113	39.0	79.0	–
	Forged rod 6-in. diam	B	108	116	8.0	12.0	–

(a) A–1800° F/1 hr plus 1325° F/8 hr, FC 100° F/1 hr to 1150° F, plus 1150° F/8 hr
B–1800° F/1 hr plus 1325° F/8 hr, FC 20° F/1 hr to 1150° F
C–1700° F/1 hr plus 1325° F/8 hr, FC 100° F/1 hr to 1150° F, plus 1150° F/8 hr
D–1700° F/1 hr plus 1325° F/8 hr, FC 20° F/1 hr to 1150° F
S–1800° F/1 hr plus 1325° F/16 hr
T–1700° F/1 hr plus 1325° F/16 hr

TABLE 53–*Summary of Tensile Results Obtained from Smooth and Sharp-Edge-Notched Specimens of Alloy 718 Cold-Worked 20 Percent and Aged*[(a)] *(ref. 52)*

Orientation	Temperature, °F	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation, percent
		Smooth Specimens		
Longitudinal	RT	218.0	207.0	9.5
Transverse	RT	217.0	203.5	7.5
Longitudinal	800	185.8	174.0	9.5
Transverse	800	190.0	178.0	9.5
Longitudinal	1000	183.3	170.0	12.0
Transverse	1000	184.6	173.0	7.0
Longitudinal	1200	174.5	145.0	10.0
Transverse	1200	173.5	158.8	8.5
		Edge-Notched Specimens		
Longitudinal	RT	219.5	–	–
Transverse	RT	204.5	–	–
Longitudinal	800	170.2	–	–
Longitudinal	800	186.5	–	–
Transverse	800	181.2	–	–
Longitudinal	1000	170.5	–	–
Transverse	1000	160.5	–	–
Longitudinal	1200	141.0	–	–
Transverse	1200	134.5	–	–

(a) 1325° F/8 hr, FC to 1150° F in 10 hr, AC

TABLE 54–*Summary of Tensile Results Obtained from Smooth and Sharp-Edge-Notched Specimens of Alloy 718, Annealed at 1750° F and Aged*[(a)] *(ref. 52)*

Orientation	Temperature, °F	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation, percent
		Smooth Specimens		
Longitudinal	RT	207.5	173.5	17.3
Transverse	RT	206.0	173.0	18.0
Longitudinal	800	182.5	155.7	20.3
Transverse	800	179.7	157.5	21.0
Longitudinal	1000	177.2	153.0	21.5
Transverse	1000	176.4	151.5	21.0
Longitudinal	1200	159.4	141.9	11.0
Transverse	1200	160.0	141.7	10.0

TABLE 54–*Summary of Tensile Results Obtained from Smooth and Sharp-Edge-Notched Specimens of Alloy 718, Annealed at 1750° F and Aged*[a] *(ref. 52)*–Concluded

Orientation	Temperature, °F	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation, percent
		Edge-Notched Specimens		
Longitudinal	RT	187.0	–	–
Transverse	RT	180.0	–	–
Longitudinal	800	136.0	–	–
Transverse	800	161.5	–	–
Longitudinal	1000	132.0	–	–
Transverse	1000	142.5	–	–
Longitudinal	1200	138.0	–	–
Transverse	1200	139.2	–	–

(a) Annealed 1 hr at 1750° F; aged 1325° F/8 hr, FC to 1150° F in 10 hr, AC

TABLE 55–*Summary of Tensile Results Obtained from Smooth and Sharp-Edge-Notched Specimens of Alloy 718, Annealed at 1950° F and Aged*[a] *(ref. 52)*

Orientation	Temperature, °F	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation, percent
		Smooth Specimens		
Longitudinal	RT	204.0	176.5	20.5
Transverse	RT	198.5	168.0	21.0
Longitudinal	800	174.2	156.0	19.3
Transverse	800	169.0	150.0	17.5
Longitudinal	1000	169.3	150.2	18.0
Transverse	1000	163.0	148.8	19.0
Longitudinal	1200	157.0	143.3	6.5
Transverse	1200	152.5	134.5	7.5
		Edge-Notched Specimens		
Longitudinal	RT	196.6	–	–
Transverse	RT	195.6	–	–
Longitudinal	800	167.5	–	–
Transverse	800	160.0	–	–
Longitudinal	1000	166.5	–	–
Transverse	1000	159.5	–	–
Longitudinal	1200	148.4	–	–
Transverse	1200	151.9	–	–

(a) Annealed 1 hr at 1950° F; aged 1350° F/8 hr, FC to 1200° F in 12 hr, AC

TABLE 56—*Effect of Mechanical and Thermal Treatments on the Short-Time Tensile Properties of 0.027-In. Alloy 718 Sheet (ref. 11)*

Prior Exposure[a] Temperature, °F	Prior Exposure Stress, 1000 psi	Prior Exposure Time, hr	Test Temperature, °F	Unnotched Tensile Strength, 1000 psi L[b]	Unnotched Tensile Strength, 1000 psi T	Yield Strength, (0.2% Offset), 1000 psi L	Yield Strength, (0.2% Offset), 1000 psi T	Elongation in 2 In., percent L	Elongation in 2 In., percent T	Sharp-Edge Notches Tensile Strength, 1000 psi L	Sharp-Edge Notches Tensile Strength, 1000 psi T	Strength Ratio, N/S[c] L	Strength Ratio, N/S[c] T
Cold-Rolled + Annealed 1800° F, 1 hr + 1325° F, 8 hr, FC 20° F/1 hr to 1150° F, then AC													
None			-110	–	212	–	168	–	25.0	–	195	–	0.92
None			RT	–	196	–	163	–	21.0	–	185	–	0.93
650	40	1000	RT	–	197	–	162	–	21.0	–	178	–	0.91
None			350	–	191	–	153	–	20.0	–	–	–	–
None			650	–	171	–	141	–	20.0	–	158	–	0.92
650	40	1000	650	–	172	–	137	–	23.0	–	163	–	0.95
None			800	–	188	–	128	–	23.0	–	156	–	0.83
None			1000	–	169	–	114	–	24.0	–	143	–	0.85
Cold-Rolled + 1325° F, 8 hr, FC 20° F/1 hr to 1150° F, then AC													
None			-320	–	260	–	229	–	13.0	–	236	–	0.91
None			-110	–	232	–	206	–	17.0	–	209	–	0.88
None			RT	–	221	–	198	–	12.0	–	196	–	0.89
650	40	1000	RT	–	212	–	195	–	13.0	–	199	–	0.94
None			350	–	205	–	188	–	12.0	–	179	–	0.88
None			650	–	193	–	179	–	12.0	–	173	–	0.90
650	40	1000	650	–	198	–	182	–	13.0	–	172	–	0.87
None			800	–	–	–	–	–	–	–	169	–	–
None			1000	–	180	–	165	–	10.0	–	166	–	0.92

(a) Conditions of exposure prior to testing
(b) L–longitudinal orientation; T–transverse orientation
(c) $N/S = \dfrac{\text{Tensile strength of notched specimen}}{\text{Tensile strength of unnotched (smooth) specimen}}$

TABLE 57—*Stress-Rupture Properties of Allvac 718 Forged Bar vs Temperature of Solution Treatment*

	Direction	Life, hr	Elongation, percent	Reduction in Area, percent
Stress Rupture at 1300° F/75,000 psi				
1750° F solution plus double age	Transverse	45.7	19.0	30.0
1800° F solution plus double age	Transverse	60.0	17.2	28.9
1800° F solution plus double age	Transverse	68.3	16.5	27.9
1950° F solution plus double age	Transverse	136.4	5.3	8.9

TABLE 57—*Stress-Rupture Properties of Allvac 718 Forged Bar vs Temperature of Solution Treatment*—Concluded

	Direction	Life, hr	Elongation, percent	Reduction in Area, percent
		Stress Rupture at 1200° F/100,000 psi		
1700° F solution plus double age	–	92.6	14.0	18.8
1800° F solution plus double age	–	160.6	15.0	19.8

TABLE 58—*Effect of Heat-Treatment Conditions on the 1000-Hr Rupture Strengths of Smooth and Notched Alloy 718 Sheet*

Stress-Concentration Factor, K_t	Temperature		
	800° F	1000° F	1200° F
	Cold-Worked and Aged		
1.0 (Smooth)	185,000 psi	125,000 psi	44,000 psi
2.3	–	–	–
6.0	–	88,000 psi	37,000 psi
>20	160,000 psi	35,000 psi(a)	25,000 psi
	Cold-Worked, Annealed at 1750° F, and Aged		
1.0 (Smooth)	170,000 psi	145,000 psi	68,000 psi
2.3	–	135,000 psi	65,000 psi
6.0	–	125,000 psi	65,000 psi
>20	140,000 psi	65,000 psi	55,000 psi
	Cold-Worked, Annealed at 1950° F, and Aged		
1.0 (Smooth)	160,000 psi	135,000 psi	75,000 psi
2.3	–	–	49,000 psi
6.0	–	87,000 psi	37,000 psi
>20	155,000 psi	60,000 psi(b)	31,000 psi(c)

(a) Transverse direction, in longitudinal direction 1000-hr strength = 65,000 psi
(b) Longitudinal direction, in transverse direction 1000-hr strength = 77,000 psi
(c) Longitudinal direction, in transverse direction 1000-hr strength = 41,000 psi

TABLE 59—*Summary of Creep and Rupture Properties of Alloy 718 as Influenced by Heat Treatments (ref. 52)*

Condition	800° F	1000° F	1200° F
Extrapolated Stress Necessary to Produce a Minimum Creep Rate of 0.000001 Percent/Hr			
Cold-worked and aged	175,000 psi	112,000 psi	10,000 psi
Annealed at 1750° F and aged	150,000 psi	110,000 psi	26,000 psi
Annealed at 1950° F and aged	155,000 psi	114,000 psi	59,000 psi
Minimum Time for Rupture of Sharp Edge-Notched Specimens Under 40,000 psi Stress			
Cold-Worked and aged	>10,000 hr	500 hr	1 hr
Annealed at 1750° F and aged	>10,000 hr	>10,000 hr	>10,000 hr
Annealed at 1950° F and aged	>10,000 hr	>10,000 hr	8 hr

TABLE 59–*Summary of Creep and Rupture Properties of Alloy 718 as Influenced by Heat Treatments (ref. 52)*–Concluded

Condition	800° F	1000° F	1200° F
	Stress for Rupture in 50,000 Hr		
Cold-worked and aged	–	98,000 psi	22,000 psi
Annealed at 1750° F and aged	–	115,000 psi	40,000 psi
Annealed at 1950° F and aged	–	110,000 psi	55,000 psi

TABLE 60–*Effect of Aluminum and Titanium on Stress-Rupture Life (1300° F/75,000 psi) of Alloy 718*

[Specimens annealed plus aged at 1325° F/8 hr, FC to 1150° F, and AC (Cb + Ta = 5.0%; B = 0.004%) (ref. 33)]

Annealing Treatment		0.15 Percent Al		0.75 Percent Al	
		0.65 Percent Ti	1.45 Percent Ti	0.65 Percent Ti	1.45 Percent Ti
1750° F/1 hr, AC	Smooth-bar life, hr	14.5	18.9	22.4	16.9
	Elongation, percent	35.0	34.0	36.0	29.0
	Notch-bar life, hr	78.2	53.5	96.9	60.0
1800° F/1 hr, AC	Smooth-bar life, hr	32.9	33.1	77.4	40.5
	Elongation, percent	25.0	31.0	17.0	24.0
	Notch-bar life, hr	308.4	189.1	258.3	156.7
1900° F/1 hr, AC	Smooth-bar life, hr	98.4	206.3	113.3	152.4
	Elongation, percent	16.0	12.0	11.0	7.0
	Notch-bar life, hr	568.6	582.4	299.6	432.4
2000° F/1 hr, AC	Smooth-bar life, hr	255.0	296.1	91.2	215.6
	Elongation, percent	8.0	8.0	8.0	4.0
	Notch-bar life, hr	247.4	385.2	129.4	20.4

TABLE 61–*Processing Techniques and Corresponding Creep-Rupture Properties of Alloy 718 Billet, Upset-Forged into Pancakes as Shown (ref. 52)*

Pancake	Heating Temp., °F	Hot Reduction, percent	Reheat Temp., °F	Hot Reduction, percent	Finish Temp., °F	Average Grain Diameter, ASTM No.	1200° F/110 ksi			Microstructure		
							Life, hr	Elong, percent	RA, percent	Ni_3Cb Platelet Size	Ni_3Cb Platelet Volume	Grain Boundary Film
A	2050	75	–	–	1840	6.5	99	8.8	11			
B	2050	75	–	–	1930	5	0.55	N[c]	–	Large	Large	Some
C	2050	42	2050	40	1820	5.1[a]	96	6.8	12.5			
D	2050	42	1950	40	1740	3.2[a]	42	N-4	10	Mixed	Large	Some
E	1950	42	2050	40	1820	4.9[a]	86	8.5	14.8			
F	1950	75	–	–	1780	8	70	14	30			
G	1950	75	–	–	1830	7.5	156	9.3	24	Medium	Medium	Little
H	1950	42	1950	40	1770	5.7[a]	180	6.3	20	Mixed	Medium	Little
I	1950	42	1950	40	1780	3.4[a]	116	N-5.5	11.5			
J	1950	71	1900	14.5	1720	3.5[a]	120	N	–	Large	Large	Little
K	2000	67	1900	25	1670	6	129	8.3	14			

TABLE 61–*Processing Techniques and Corresponding Creep-Rupture Properties of Alloy 718 Billet, Upset-Forged into Pancakes as Shown (ref. 52)*–Concluded

Pancake	Heating Temp., °F	Hot Reduction, percent	Reheat Temp., °F	Hot Reduction, percent	Finish Temp., °F	Average Grain Diameter, ASTM No.	1200° F/110 ksi Life, hr	Elong, percent	RA, percent	Microstructure Ni$_3$Cb Platelet Size	Ni$_3$Cb Platelet Volume	Grain Boundary Film
L	1850	75	–	–	1680	>8	77	17.5	33	Small	Large	Little
M	1950	50	1900	50	1640	7.8(a)	68	15.5	25.5			
As Rec'd							215	13				
Upset 3 to 1							201	6.5				

(a) Weighted average of duplex grain structure
(b) Average of duplicate tests heat treated 1750° F/1 hr, AC + 1325° F/8 hr, FC at 100° F/1 hr to 1150° F/8 hr, AC; 0.375-in.-diam specimens were machined from radial direction
(c) Notch failure

TABLE 62–*High-Temperature Tensile Properties as a Function of Hydrostatic Extrusion and and Aging Temperature of Wrought Alloy 718 (ref. 44)*

Pre-Extrusion Treatment	Amount of Cold Work	Post-Aging Treatment	Yield Strength (0.2% Offset), ksi	Ultimate Tensile Strength, ksi	Elongation, percent	Reduction in Area, percent
1	0	None	136	156	12	15
2	50%	950° F/8 hr, AC	150.7	182	19	19
2	50%	1050° F/8 hr, AC	158	194	15	19
2	50%	1150° F/8 hr, AC	162	185	15	20
2	50%	1250° F/8 hr, AC	168	187	14	21

TABLE 63–*Time Requirements for Heating and Quenching Various Thicknesses of René 41 Material (ref. 64)*

Material Thickness, in.	Holding Time, min	Maximum Quench Delay, sec
Annealing (2025° ± 25° F) WQ		
0.039 and under	4 ± 1	5
0.040 - 0.099	6 ± 1	5
0.100 - 0.249	12 ± 2	5
0.250 - 0.499	20 ± 4	7
0.500 - 0.749	30 ± 5	7
0.750 - 1.000	40 ± 5	7
Solution Treatment (1975° ± 25° F) WQ		
0.059 and under	10 ± 2	5
0.060 - 0.249	15 ± 3	5
0.250 - 0.449	20 ± 4	7
0.500 - 0.999	30 ± 5	7
1.00 - 1.99	50 ± 5	7
2.00 - 3.00	70 ± 5	7

TABLE 64–*Effect of Heat Treatment on Stress-Rupture Properties of R-41 Bar (ref. 7)*

Condition	Test Temperature, °F	Stress, 1000 psi, for Rupture in		
		10 Hr	100 Hr	1000 Hr
Solution heat-treated	1200	–	–	102
2 hr at 2150° F, AC;	1300	–	95	80
aged 4 hr at	1400	92	68	50
1650° F, AC	1500	65	45	29
	1600	45	28	17
	1700	28	18	11
Solution heat-treated	1200	–	–	100
4 hr at 1950° F, AC;	1300	–	96	74
aged 16 hr at	1400	90	64	40
1400° F, AC	1500	60	38	24
	1600	37	23	14
	1700	23	12	–

TABLE 65–*Room-Temperature Tensile Properties of Solution-Annealed René 41 Foil as Affected by Aging in Air*

Condition	Yield Strength (0.2% Offset) ksi	Ultimate Tensile Strength, ksi	Elongation in 2 in., percent	Modulus of Elasticity, 10^6 psi
ST - 2150° F/1 hr, AC:				
ST + 1650° F/4 hr, AC	102	116	4	27
ST + 1600° F/5 hr, AC	102	116	4	27
ST + 1600° F/10 hr, AC	104	119	3	29
ST + 1900° F/5 hr, AC	62	96	12	26
ST + 1900° F/10 hr, AC	48	78	10	26

TABLE 66–*Effect of Hydrostatic Extrusion* on the Room-Temperature Properties of Wrought René 41*

Condition	Yield Strength (0.2% Offset), ksi	Ultimate Tensile Strength, ksi	Elongation, percent	Reduction in Area, percent	Hardness, R_C
1. 1975° F/4 hr, AC + 1400° F/16 hr, AC	133.8	193.2	7	12	37
2. 1975° F/4 hr, AC + 1400° F/16 hr, AC + 50% hydrostatic extrusion	243.0	291.0	3	11	50
3. 1975° F/4 hr, AC + 1400° F/16 hr, AC + 50% hydrostatic extrusion + 1400° F/16 hr AC	280.0	280.0	0	0	56

*Extruded at room temperature

TABLE 67–*Room-Temperature Compression Properties of Cast René 41 Bars Showing the Influence of Aging at 1400° F*

Condition	Compression Yield (0.2% Offset), psi	Compression* Ultimate, psi	Modulus of Elasticity 10^6 psi
1950° F/4 hr, AC	82,500	375,000	6.4
1950° F/4 hr, AC	82,500	440,000	6.9
1950° F/4 hr, AC + 1400° F/16 hr, AC	97,100	348,000	7.8
1950° F/4 hr, AC + 1400° F/16 hr, AC	97,500	336,000	7.0

*Load at failure based on original area

TABLE 68–*Impact Strength of Cast René 41*

Condition	Test Temperature, °F	Charpy V Notch Impact Strength: Impact Strength, ft-lb
1950° F/4 hr, AC + 1400° F/16 hr, AC	-40	7.2
1950° F/4 hr, AC + 1400° F/16 hr, AC	-40	7.9
1950° F/4 hr, AC + 1400° F/16 hr, AC	-40	10.0
1950° F/4 hr, AC + 1400° F/16 hr, AC	-40	7.9
1950° F/4 hr, AC + 1400° F/16 hr, AC	RT	10.4
1950° F/4 hr, AC + 1400° F/16 hr, AC	RT	6.9
1950° F/4 hr, AC + 1400° F/16 hr, AC	RT	7.3
1950° F/4 hr, AC + 1400° F/16 hr, AC	RT	8.1
2150° F/2 hr, AC + 1650° F/4 hr, AC	RT	7.0
2150° F/2 hr, AC + 1650° F/4 hr, AC	RT	10.8
1950° F/4 hr, AC + 1400° F/16 hr, AC	1200	8.2
1950° F/4 hr, AC + 1400° F/16 hr, AC	1200	8.1
1950° F/4 hr, AC + 1400° F/16 hr, AC	1200	7.4
1950° F/4 hr, AC + 1400° F/16 hr, AC	1200	6.1

TABLE 69–*Elevated-Temperature Tensile Properties of René 41 0.040-In. Sheet–Comparison of Single and Double Age*[a]

Test Temperature, °F	Ultimate Tensile Strength, 1000 psi		Yield Strength (0.2% Offset) 1000 psi		Elongation in 2 In., percent	
	Single	Double	Single	Double	Single	Double
70	204	204	156	151	21	21
800	175	186	135	141	17	10
1200	172	175	127	125	12	10
1400	143	151	118	124	4.5	7.5
1600	96	91	90	81	6.5	12
1800	39	39	35	32	16	19

(a) Single age–1400° F/16 hr; double age–1650° F/1 hr plus 1400° F/10 hr

TABLE 70—*Creep-Rupture Properties of Cast René 41 Bars*

Condition	Test Temp, °F	Stress, psi	Life, hr	Minimum Creep Rate, %/hr	Elongation, percent	Reduction of Area, percent
2150° F/2 hr, AC + 1650° F/4 hr, AC	1650	25,000	72.8	0.145	6.0	7.8
2150° F/2 hr, AC + 1650° F/4 hr, AC	1600	40,000	31.7	0.028	6.0	7.0
2150° F/2 hr, AC + 1650° F/4 hr, AC	1400	60,000	101.7	0.007	1.0	0.8
1950° F/4 hr, AC + 1400° F/16 hr, AC	1400	60,000	149.5	0.006	4.0	5.5

TABLE 71—*Short-Time Tensile Properties at -110° to 1200° F of Cold-Rolled and Aged 0.030-In. René 41 Sheet*

[Effect of degree of cold work and prior exposure at elevated temperature and stress]

Prior Exposure(a)				Unnotched (Smooth) Specimens						Sharp-Edge Notches			
Temperature, °F	Stress, 1000 psi	Time, hr	Test Temperature °F	Tensile Strength, 1000 psi		Yield Strength (0.2% Offset), 1000 psi		Elongation in 2 In., percent		Tensile Strength, 1000 psi		Strength Ratio, N/S	
				L(b)	T	L	T	L	T	L	T	L	T
Cold-Reduced 20 Percent Plus 16 Hr 1400° F													
None			-110	247	244	214	209	14	14	221	190	0.90	0.78
None			RT	229	226	208	200	10	10	205	181	0.89	0.80
650	40	1000	RT	230	–	205	–	12	–	196	180	0.85	–
None			650	215	209	190	188	11	12	160	132	0.74	0.63
650	40	1000	650	213	208	189	185	11	11	160	166	0.75	0.80
None			800	213	208	192	185	13	9	126	130	0.59	0.62
Cold-Reduced 35 Percent Plus 2 Hr, 1500° F													
None			-110	268	255	246	229	10	10	199	199	0.74	0.78
650	40	1000	-110	268	254	244	229	10	9	–	184	–	0.72
1000	40	1000	-110	269	258	244	234	10	5	190	–	0.71	–
None			RT	249	237	230	216	8	7	196	182	0.79	0.77
650	40	1000	RT	246	242	230	222	9	7	182	190	0.74	0.78
1000	40	1000	RT	251	240	237	221	7	6	147	161	0.59	0.67
None			350	243	–	217	–	9	–	181	174	0.74	–
None			650	229	222	210	200	7	8	138	155	0.60	0.70
650	40	1000	650	229	221	213	200	6	7	160	158	0.70	0.72
1000	40	1000	650	230	228	219	218	7	7	140	143	0.61	0.63
None			800	221	217	210	196	6	6	142	142	0.64	0.65
None			1000	221	218	201	194	3.8	5	116	140	0.53	0.64
1000	40	1000	1000	227	219	206	193	3.0	3.5	133	116	0.59	0.53
None			1200	228	223	189	183	15	10	119	115	0.52	0.52

(a) Conditions of exposure prior to testing
(b) L–longitudinal orientation; T–transverse orientation

TABLE 72—*Tensile Properties of Age-Hardened René 41 Sheet at Cryogenic Temperatures*
[0.020-in. sheet solution treated at 1950° F/30 min, AC, and aged 1400° F/16 hr]

Test Temperature, °F	Direction	Yield Strength, 1000 psi	Tensile Strength, 1000 psi	Elongation percent	Notched-Unnotched Ratio, $K_t = 6.3$
78	Longitudinal	138	181	18	0.91
	Transverse	134	174	12	0.94
-100	Longitudinal	148	192	13	0.90
	Transverse	145	183	10	0.96
-320	Longitudinal	161	202	9	0.94
	Transverse	162	196	7	0.95
-423	Longitudinal	179	212	6	0.99
	Transverse	174	206	5	1.00

TABLE 73—*Effect of Solution-Treatment Temperature on Tensile Properties of Waspaloy (ref. 9)*

Test Temperature, °F	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 2 In., percent	Reduction in Area, percent
	1975° F/4 Hr, AC, Plus 1550° F/24 Hr, AC, Plus 1400° F/16 Hr, AC			
70	190	130	24	24
1000	172	125	16	20
1200	165	105	23	32
1400	125	99	28	41
1600	80	76	30	54
	1825° F/4 Hr, OQ, Plus 1550° F/4 Hr, AC, Plus 1400° F/16 Hr, AC			
70	200	140	20	22
1000	172	125	18	22

TABLE 74—*Room-Temperature Tensile Properties of Waspaloy Sheet*
[Effect of cold-working and aging treatments]

Heat Treatment	Direction[a]	Condition of Material Prior to Aging				
		Ultimate Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation in 2 In., percent	NS[b] 1000 psi	N/S
		Annealed				
None	L	129	62	64	87	0.67
	T	129	61	63	91	0.70
16 hr at 1300° F	L	179	127	34	148	0.83
	T	183	125	36	151	0.82
2 hr at 1400° F	L	181	127	34	140	0.77
	T	179	123	36	141	0.79
16 hr at 1400° F	L	191	139	31	154	0.81
	T	190	135	31	151	0.80
2 hr at 1500°	L	190	136	32	154	0.81
	T	187	132	33	154	0.82
24 hr at 1550° F plus 16 hr at 1400° F	L	186	120	28	143	0.77
	T	190	128	27	147	0.77

TABLE 74–*Room-Temperature Tensile Properties of Waspaloy Sheet*–Concluded
[Effect of cold-working and aging treatments]

Heat Treatment	Direction(a)	Condition of Material Prior to Aging: Ultimate Tensile Strength, 1000 psi	Yield Strength, 1000 psi	Elongation in 2. In., percent	NS(b) 1000 psi	N/S
		Cold-Worked 20 Percent				
None	L	181	124	25	149	0.82
	T	152	121	25	152	1.00
16 hr at 1300° F	L	218	190	15	197	0.90
	T	213	183	13	201	0.94
2 hr at 1400° F	L	207	179	14	197	0.95
	T	211	176	15	194	0.92
16 hr at 1400° F	L	–	–	–	–	–
	T	–	–	–	–	–
2 hr at 1500° F	L	205	176	16	191	0.93
	T	204	174	18	202	0.99
24 hr at 1500° F plus 16 hr at 1400° F	L	–	–	–	–	–
	T	–	–	–	–	–
		Cold-Worked 40 Percent				
None	L	189	171	5	187	0.99
	T	186	162	8	190	1.02
16 hr at 1300° F	L	243	219	6	212	0.87
	T	240	215	7	198	0.82
2 hr at 1400° F	L	241	219	6	195	0.81
	T	232	205	8	187	0.81
16 hr at 1400° F	L	–	–	–	–	–
	T	–	–	–	–	–
2 hr at 1500° F	L	237	216	9	212	0.89
	T	232	206	11	201	0.87
24 hr at 1550° F plus 16 hr at 1400° F	L	–	–	–	–	–
	T	–	–	–	–	–

(a) L = longitudinal
T = transverse
(b) NS–tensile strength of sharp-edge notch sample

TABLE 75–*Short-Time Tensile Properties of 0.030-In. Waspaloy Sheet After Various Treatments*

Prior Exposure(a): Temperature, °F	Stress, 1000 psi	Time, hr	Test Temperature, °F	Unnotched (Smooth) Specimens: Tensile Strength, 1000 psi L(b)	Tensile Strength, 1000 psi T(b)	Yield Strength, 1000 psi L	Yield Strength, 1000 psi T	Elongation in 2 In., percent L	Elongation in 2 In., percent T	Sharp-Edge Notches: Tensile Strength, 1000 psi L	Tensile Strength, 1000 psi T	Strength Ratio N/S L	Strength Ratio N/S T
				Annealed Plus Aged 16 Hr at 1400° F									
None			-110	207	201	145	139	39	33	151	163	0.73	0.81

TABLE 75–*Short-Time Tensile Properties of 0.030-In. Waspaloy Sheet After Various Treatments*–Concluded

Prior Exposure[a]				Unnotched (Smooth) Specimens						Sharp-Edge Notches			
Temperature, °F	Stress, 1000 psi	Time, hr	Test Temperature, °F	Tensile Strength, 1000 psi		Yield Strength, 1000 psi		Elongation in 2 In., percent		Tensile Strength, 1000 psi		Strength Ratio N/S	
				L[b]	T[b]	L	T	L	T	L	T	L	T
				Annealed Plus Aged 16 Hr at 1400° F									
None			RT	191	190	139	135	31	31	154	151	0.81	0.80
650	40	1000	RT	186	184	135	132	31	31	161	163	0.87	0.89
1000	40	1000	RT	192	187	144	137	31	30	159	159	0.83	0.85
None			650	167	167	122	124	33	31	146	139	0.87	0.83
650	40	1000	650	166	163	124	118	31	30	138	136	0.83	0.83
None			800	162	163	120	120	32	34	136	139	0.84	0.85
None			1000	158	158	119	119	33	27	139	131	0.88	0.83
1000	40	1000	1000	163	162	129	126	29	29	138	140	0.85	0.86
				Cold-Reduced 40 Percent Plus Aged 2 Hr at 1500° F									
None			-110	250	240	221	210	15.0	14	216	203	0.86	0.85
650	40	1000	-110	253	239	224	208	16.0	17	208	209	0.82	0.87
1000	40	1000	-110	255	244	223	214	11.0	13	192	186	0.75	0.76
None			RT	237	232	216	206	9.0	11	212	201	0.89	0.87
650	40	1000	RT	227	224	206	201	10.0	9	196	196	0.86	0.88
1000	40	1000	RT	234	228	215	208	9.0	9	193	186	0.83	0.82
None			650	211	204	193	184	9.0	9	163	149	0.77	0.73
650	40	1000	650	246	204	228	184	8.0	10	176	161	0.72	0.79
1000	40	1000	650	219	209	198	186	7.0	8	158	160	0.72	0.77
None			800	204	199	188	180	7.0	7	172	166	0.84	0.83
None			1000	204	199	177	176	4.5	7	169	166	0.83	0.83
1000	40	1000	1000	213	214	190	180	3.5	6	122	114	0.57	0.53

(a) Conditions of exposure, if any, between heat treatment and testing
(b) L–longitudinal orientation; T–transverse orientation

TABLE 76–*Rupture and Creep Properties of 0.030-In. Waspaloy Sheet*

	Temperature		
	800° F	1000° F	1200° F
Stress for Rupture in 50,000 Hr, 1000 psi			
Cold reduced and aged[a]	–	122	52
Annealed and aged[b]	–	110	–
Stress for a Minimum Creep Rate of 0.000001 Percent/Hr, 1000 psi			
Cold reduced and aged[a]	164	124	47
Annealed and aged[b]	–	112	66

(a) Cold reduced 40%, aged 2 hr at 1500° F
(b) Aged 16 hr at 1400° F

TABLE 77–*Estimated Minimum Time for Rupture of Notched Specimens Under a Net Section Stress of 40,000 psi*

Direction	Temperature	
	1000° F	1200° F
Cold-Reduced and Aged[a]		
Longitudinal	9500 hr	4000 hr
Transverse	700 hr	100 hr
Annealed and Aged[b]		
Longitudinal	6000 hr	120 hr
Transverse	6000 hr	120 hr

(a) Cold-reduced 40%, aged 2 hr at 1500° F
(b) Aged 16 hr at 1400° F

TABLE 78–*Tensile Properties of Waspaloy Bar Stock at Cryogenic Temperatures (ref. 26)*

Test Temperature, °F	Yield Strength, 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Reduction in Area, percent	Notched Strength, 1000 psi
78	122	195	28	28	215
-104	132	203	24	26	234
-320	157	235	17	18	250
-423	163	243	18	14	262

Note: Specimen test section 1/4-in. diam with 1-in. gage length. Each notched specimen had 60-deg V-notch 0.037 in. deep with 0.005-in. notch radius.

TABLE 79–*Test Results Obtained by Ladish Co. (ref. 75)*

Processing	Yield Strength (0.2% Offset), ksi	Tensile Strength, ksi	Elongation, percent	Reduction of Area, percent
Standard forging procedure	137	166	9	12
Thermomechanical forging procedure	145	195	19	21

TABLE 80–*Tensile Properties of Udimet 500 Wrought Bar*[a] *(ref. 5)*

Temperature, °F	Tensile Strength, 1000 psi	Elongation in 2 In., percent	Reduction in Area, percent
RT	188	15.0	12.5
1200	175	18.0	18.0
1300	167	19.0	20.0
1400	156	21.0	23.0
1500	144	21.5	27.0
1600	100	21.5	32.0
1700	79	22.0	36.0
1800	45	22.0	40.0

(a) Heat treatment: 1975° F/4 hr, AC; 1550° F/24 hr, AC; 1400° F/16 hr, AC

TABLE 81–*Stress-Rupture Strength of Udimet 500 Wrought Bar*[a] *(ref. 5)*

Temperature, °F	Stress, 1000 psi for Rupture in		
	10 hr	100 hr	1000 hr
1350	96	78	61
1400	86	73	51
1500	65	48	33
1600	47	32	21
1700	31	20	12
1800	20	12	–

(a) Heat treatment: 2150° F/2 hr, AC; 1975° F/4 hr, AC; 1550° F/24 hr, AC; 1400° F/16 hr, AC

TABLE 82–*Effect of Eliminating Intermediate Aging on Room-Temperature Properties of Udimet 500 (ref. 15)*

Condition	Tensile Strength[a], 1000 psi	Yield Strength[a], 1000 psi	Elongation[a], percent	Reduction in Area[a], percent
1[b]	145	119	5.5	7.9
2[c]	174	119	14.3	16.7

(a) Average of two tests
(b) 1975° F/4 hr, AC; 1550° F/24 hr, AC; 1400° F/16 hr, AC
(c) 1975° F/4 hr, AC; 1400° F/16 hr, AC

TABLE 83—*Room-Temperature Hardness Data for Hastelloy C in Various Forms and Conditions*

Form	Condition	Aging		Hardness, Rockwell
		Temperature, °F	Time, hr	
Sheet, 0.050 in.-thick	Heat-treated at 2225° F, RAC	–	–	B-91
	Heat-treated at 2225° F, RAC and aged	1100	16	B-98
		1200	16	B-102
		1200	48	B-105
		1200	72-1/2	B-103
Bar	Heat-treated at 2225° F, RAC	–	–	B-94
Sand cast	Heat-treated 2250° F, RAC	–	–	B-93
Investment cast	As-cast	–	–	B-96
	As-cast and aged	1475	5	C-25.0
		1475	25	C-37.5
		1475	100	C-41.0
		1475	1000	C-39.0

TABLE 84—*Effect of Heat Treatment on Room-Temperature Tensile Properties of Hastelloy C*

Form	Condition	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 2 In., percent	Reduction in Area, percent
Sheet, 0.050-in. thick	Heat-treated at 2225° F, RAC	128	68	49.0	–
	Heat-treated at 2225° F, RAC and aged:				
	16 hr at 1100° F	145	82	44.5	–
	16 hr at 1200° F	126	67	47.8	–
	48 hr at 1200° F	132	72	22.0	–
	72-1/2 hr at 1200° F	135	75	19.3	–
	16 hr at 1600° F	132	76	10.0	–
Investment cast	As-cast and aged at 1475° F for:				
	5 hr	80	62	7.0	12.8
	25 hr	98	76	4.0	4.9
	50 hr	109	90	1.0	1.5
	100 hr	113	98	0.0	2.5
	500 hr	122	105	0.0	0.0
	1000 hr	122	97	2.0	1.7
	As-cast and aged at 1600° F for 16 hr	109	102	1.8	–

TABLE 85—*Effect of Solution Annealing at 2050° F (Instead of at 2225° F) on Short-Time Tensile Properties of 0.250-In. Hastelloy C Plate (ref. 70)*

Heat Treatment	Test Temperature, °F	Yield Strength (0.2% Offset), 1000 psi	Ultimate Tensile Strength, 1000 psi	Elongation in 2 In., percent
As-received	RT	56	116	60.0
2050° F/2 hr, AC	RT	53	124	34.5
As-received	1400	38	82	58.0
2050° F/2 hr, AC	1400	35	90	39.5

TABLE 86—*Effect of Cold Working on the Room-Temperature Tensile Properties of Hastelloy C (ref. 4)*

Form		Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 2 In., percent
	Cold-drawn, percent			
Bar, 3/4-in.[a]	10	134	104	44.0[b]
	20	167	166	16.5[b]
	30	146	134	29.0[b]
	40	191	191	11.5[b]
	Cold-reduced, percent			
Wire, 0.241-in. original diam[a]	0	115	–	50.0
	5	134	98	30.0
	10	139	110	32.0
	15	151	130	21.8
	20	164	133	17.0
	25	174	155	13.0
	30	185	165	4.3
	35	198	191	6.0
	40	204	197	6.0

(a) Developmental data
(b) Elongation in 1 in.

TABLE 87—*Typical Short-Time Tensile Data for Hastelloy C Sheet*

Condition	Test Temperature, °F	Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Proportional Limit, 1000 psi	Elongation in 2 In., percent
Sheet, 0.109-in.	RT	121	57	26	47.5
Heat-treated at 2225° F, WQ	1000	99	43	31	52
	1200	97	43	32	55
	1400	78	40	31	59
	1600	56	37	28	47
	1800	31	18	13	49
	2000	18	9	6	36
Sheet, 0.088-in.	RT	143	113	–	24
Cold-reduced 20%	RT	145	118	–	23
	600	123	91	–	26
	600	126	104	–	24.5

TABLE 88—*Typical Short-Time Tensile Properties and Effect of Heat Treatment on Cast Hastelloy C (ref. 4)*

Form	Condition	Test Temperature °F	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Elongation in 1 In., percent	Reduction in Area, percent
Sandcast	Heat-treated 1 hr 2250°F, RAC	RT	83	50	9.5[a]	12.0
		500	77	34	20.0	–
		750	73	34	20.2	–
		1000	64	34	13.4	–
		1200	58	31	14.4	–
		1500	54	–	18.2[a]	–
		1750	30	–	14.0[a]	–
		1950	13	–	27.5[a]	–
Investment cast	As-cast	RT	89	52	11.0	12.2
		600	70	36	7.3	7.5
		1000	67	–	12.5	14.9
		1200	62	–	15.5	15.7
		1500	56	–	18.2	15.4
		1750	32	–	14.0	49.0
		1950	15	–	27.5	60.0
	As-cast and aged at 1475°F for:					
	25 hr	1500	61	–	15.0	21.8
	50 hr	1500	67	–	19.0	25.7
	100 hr	1500	60	–	22.0	44.1
	500 hr	1500	63	–	12.0	34.6
	1000 hr	1500	64	–	6.0	39.9

(a) Elongation in 2 in.

TABLE 89—*Average Stress Rupture Data for Hastelloy Alloy C in Various Forms and Conditions*

Form	Condition	Test Temperature, °F	Average Initial Stress, 1000 psi for Rupture at				
			10 hr	50 hr	100 hr	500 hr	1000 hr
Sheet, 0.050- to 0.141-in. thick	Heat-treated at 2225°F, RAC	1200	69	57	54	47	44
		1350	50	37	33	26	23
		1500	24	20	18	14	12
		1600	17	–	10	–	6
Bar, 1- to 3.5-in. diam	Heat-treated at 2225°F, RAC	1200	69	58	55	47	43
		1350	48	37	33	25	23
		1500	26	20	18	14	12
Investment cast	As-cast	1200	–	–	49	–	42
		1350	–	–	32	–	25
		1500	25	–	18	–	14
	As-cast and aged 16 hr at 1600°F	1500	27	–	21	–	15
	As-cast	1600	–	–	13	–	9
		1700	–	–	9	–	6

TABLE 90—*Short-Time Tensile Properties of As-Cast Alloy 713C*

Test Temperature, °F	Yield Strength, (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation, percent	Reduction in Area, percent
70	106	123	7.9	11.6
1000	102	125	9.7	17.0
1200	104	125	6.7	10.5
1400	108	136	5.9	10.5
1500	95	120	6.0	11.5
1600	72	105	13.9	20.0
1700	55	85	11.8	17.7
1800	44	68	19.7	25.0

TABLE 91—*Creep-Rupture Properties of As-Cast Alloy 713C*

Test Temperature, °F	Stress, 1000 psi, for Rupture in			
	10 hr	100 hr	1000 hr	10,000 hr
1350	–	(97)[a]	76	56
1500	(86)[a]	60	44	30
1700	(42)[a]	30	18	12
1800	29	21	13	–
2000	9	6	–	–

(a) Denotes extrapolated values

TABLE 92—*Effect of Heat Treatment on Typical Stress-Rupture Properties of Vacuum-Melted, Vacuum-Cast Alloy 713C*

Condition	Temperature, °F	Stress, 1000 psi	Life, hr	Elongation, percent
As-cast	1700	30	76	7
2150°F/2 hr, AC	1700	30	121	4
2150°F/2 hr, AC plus 1700°F/16 hr, AC	1700	30	131	5
As-cast	1350	90	232	5
2150°F/2 hr, AC	1350	90	28	2
2150°F/2 hr, AC plus 1700°F/16 hr, AC	1350	90	274	5

TABLE 93—*Tensile Properties of Alloy 713LC Specimens Machined from Rotor Hubs (ref. 12)*

Condition	Yield Strength (0.2% Offset), 1000 psi	Tensile Strength, 1000 psi	Elongation in 2 In., percent	Reduction in Area, percent	Remarks
As-cast	96	101	17.0	20.0	Material containing less
2150°F/2 hr plus AC	121	140	10.5	17.5	than 0.5% Fe
As-cast	89	94	16.0	–	Material containing
2150°F/2 hr plus AC	101	112	18.2	–	about 1.5% Fe

TABLE 94–*Effect of Solution-Treatment Conditions on the 1300°F Tensile Properties of IN-100*

Condition	Ultimate Tensile Strength, 1000 psi	Yield Strength (0.2% Offset), 1000 psi	Yield Strength (0.02% Offset), 1000 psi	Elongation, percent	Reduction in Area, percent
As-cast	155	127	111	7.0	9.3
	159	121	102	12.5	19.7
	154	125	102	7.0	13.9
2150°F/2 hr/RAC	136	120	103	5.0	7.0
	135	120	96	5.0	4.8
	143	117	98	8.0	9.2
2050°F/24 hr/RAC	132	117	97	4.5	6.3
	132	115	101	5.0	10.1
	124	112	88	4.0	7.8
1900°F/24 hr/RAC	139	121	87	3.5	4.0
	136	120	105	3.5	8.6
	132	117	93	4.0	4.7

TABLE 95–*Effect of Solution-Treatment Conditions on the Elevated-Temperature Rupture Life of IN-100*

[Cast bars tested at 29,000 psi and 1800°F]

Condition	Life, hr	Elongation, percent	Reduction in Area, percent
As-cast	25.4	6.4	8.0
	36.7	7.3	10.5
	33.4	7.1	7.9
2150°F/2 hr/RAC	33.4	4.3	8.8
	32.6	8.7	11.1
	29.9	7.3	12.1
1900°F/24 hr/RAC	21.3	8.7	8.4
	30.8	9.3	16.8
	32.7	7.6	9.2
2050°F/24 hr/RAC	28.1	7.4	10.0
	33.4	8.9	10.8
	16.2	5.1	7.1

TABLE 96–*Overaging Conditions to Promote Structural Changes of MAR-M 200 (ref. 77)*

Heat Treatment	Structural Change
1. 1700°F for 512 hr	$M_{23}C_6$ precipitation
2. 2000°F for 24 hr	γ' agglomeration and small amount of M_6C
3. 2000°F for 512 hr	M_6C precipitation + γ' agglomeration
4. 2000°F for 512 hr + 1700°F for 512 hr	M_6C + $M_{23}C_6$ precipitation + γ' agglomeration

TABLE 97–*Stress-Rupture Properties at 1400° F of Single-Crystal MAR-M 200 in the As-Cast and the Heat-Treated Conditions (ref. 78)*

Material	As-Cast			Heat-Treated 2250° F/1 hr, 1600° F/32 hr		
Test Condition	1400° F/100,000 psi			1400° F/100,000 psi		
Test Direction	[001]	[100]	[010]	[001]	[100]	[010]
Rupture Life, hr	325	71	82	459	365	443
Reduction of Area, %	12	12	8	14	11	11

TABLE 98–*Tensile-Test Data of TAZ-8 Alloy Containing 1.8% V (Shown as Isolated Points in Figure 117) (ref. 90)*

Condition	Temperature, °F	Ultimate Tensile Strength, psi	Elongation, percent
As-cast	Room	133,700	1.2
		126,400	9.5
		143,500	4.8
	1800	84,500	7.2
		75,700	4.8
	1900	54,100	11.9
		54,400	7.2
		58,500	7.6
	2000	50,900	7.2
		47,500	7.2
	2100	29,700	26.8
		39,000	6.9
As-forged	Room	173,400	3.0
		165,000	3.0
		136,600	0
	1800	61,000	1.5
		56,000	2.0
	1900	49,100	6.0
		48,700	6.0
	2000	42,400	4.0
		37,300	3.0
	2100	20,600	6.0
		15,800	10.0

TABLE 99–*Tensile Properties of Hot-Rolled TAZ-8 Alloys Showing the Effect of Heat Treatment and Composition*

Melt No.*	Heat Treatment	Ultimate Tensile Strength, ksi			Elongation, percent	
		Room Temperature		1900° F	Room Temperature	1900° F
		YS	UTS			
	As-rolled	–	–	37.7	–	5.0
385, 425	+ 2200° F/3 hr, AC	–	128.3	36.4	2	8
385, 425	+ 2200° F/3 hr, OQ	149.2	162.1	33.2	2	6
425	+ 2200° F/24 hr, AC	138.6	164.5	34.5	3	3
385	+ 2200° F/64 hr, AC	–	144.8	–	6.0	–
425, 385	+ 2250° F/3 hr, AC	–	136.5	36.5	0.0	1.0
425, 424	2150° F/64 hr + 1450° F/64 hr	207	217	50.6**	0.0	2.0
424,	2150° F/61 hr + 1600° F/24 hr	189.4	226.6	57.7	2.0	4.5
425	2200° F/64 hr + 1900° F/24 hr	123.3	144.9	34.3	–	10

*

Heat	Cr	Al	W	Mo	V	Zr	C	Ta
385	5.8	6.0	3.9	4	1.8	1.1	.12	7.75
424	7	6.1	4	4	2.0	1.0	.17	8.0
425	7.9	6.0	3.8	3.9	1.63	.87	1.73	7.8

**Oxidized.

CHAPTER 3

Conclusions and Recommendations

It is clear that nickel and nickel-base alloys respond to a wide variety of thermal and mechanical treatments which profoundly influence mechanical properties. As a result, a tremendous range of mechanical properties is available in these materials. It is equally evident that a vast amount of information on the performance and the properties of nickel and nickel-base alloys has been accumulated. Indeed, it can be said that nickel metallurgy is a well developed technology.

In spite of the very considerable understanding of nickel and its alloys which has been achieved, much more research is needed. This is to be expected even of the most advanced and intensely researched fields of science and technology. Thus, nickel and its alloys continue to present untapped potential and to offer a challenging spectrum of opportunities. Examples relating to thermal and mechanical treatment are given in the following paragraphs.

Despite the observation that forming wrought nickel alloys at cryogenic temperatures does not increase strength properties significantly, there may nevertheless be advantages to carrying out forming operations at such temperatures. It is quite possible that the low temperature would tend to impede localized slip during plastic deformation, thus, forcing an increased number of slip planes and slip systems over a comparatively large region to participate. The result would be increased uniform elongation and, consequently, increased forming capability. Thus, it is possible that a component which is difficult to form at room temperature might be easier to form at cryogenic temperatures.

It should be pointed out that relatively little work has been done on the "thermomechanical treatment" of nickel-base alloys. Although interest has been shown (refs. 35, and 93 to 100) in this field, only a small amount of information has been published on the alloys discussed in this report. In contrast, the U.S.S.R. appears to be quite active in this area. Thermomechanical treatment appears to offer potential for improved properties of many alloy systems and, therefore, it is recommended that much more research be conducted.

Recent efforts to improve the strength of nickel-base superalloys (above 1600° F) have followed the trend of utilizing cast parts rather than heat treatment or thermomechanical processing of wrought parts. In this respect, the heat treatment of superalloy castings has been found to improve both strength and ductility in certain instances.

Bright annealing and bright hardening of nickel-base alloys containing chromium, aluminum, or titanium are difficult to accomplish. A vacuum or an inert gas atmosphere that is extremely pure and dry is usually required. Otherwise, a stable tenacious oxide film will form which can be removed only with great difficulty. It would be possible to simplify these operations and extend their range of applicability if a suitable coating were developed to protect the metal during these thermal treatments.

Much effort has been directed toward developing coatings to protect steel during heat treatment, but all have some shortcomings. A satisfactory coating for nickel alloys should be capable of application by a simple method such as dipping, spraying, or brushing; moreover, it should crack or spall on quenching, or be otherwise easily removed after the heat-treating cycle has been completed.

Grain boundaries influence strength by impeding dislocation movement. Thus, fine-grain materials with their greater proportion of grain boundaries per unit volume exhibit greater room-temperature strength than the same material in the coarse-grain condition. At elevated temperatures, where the deformation mechanism is different, the reverse is actually the case. Less grain boundary area, particularly perpendicular to the applied stress, has been found to enhance high-temperature properties. Therefore, it was not unexpected when unidirectional solidification and single-crystal parts provided improved strength at elevated temperatures.

To change the grain size of nickel and wrought nickel alloys requires controlled hot working or cold working followed by annealing. It may be possible to so regulate these operations that extremely fine grains are produced, perhaps of submicroscopic size. Nickel-base alloys of extremely fine grain could well have unusual strength properties; this could be especially true of those alloys which respond strongly to age hardening.

It is recommended that further consideration be given to the above items, some or all of which may deserve additional research.

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