DMIC Report 217
June 1, 1965

PHYSICAL METALLURGY OF ALLOY 718

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Roger J. Runck

Roger J. Runck
Director

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by

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to

OFFICE OF THE DIRECTOR OF DEFENSE
RESEARCH AND ENGINEERING

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by

H. J. Wagner and A. M. Hall*

SUMMARY

Wrought, weldable nickel-chromium-base Alloy 718 was introduced about 5 years ago for service at medium temperatures, i.e., to about 1300 F. Though not as strong as the nickel-base superalloys in the 1500 to 1800 F range, its combination of good mechanical and fabrication properties at both high and low temperatures has earned it an important role in a number of aerospace systems. It has been used for its good cryogenic properties in cryogenic tankage for rockets; its short-time strength at temperatures to 1200 F has permitted its use in liquid-fueled rocket engines; its creep-rupture properties at temperatures up to 1300 F have enabled it to be used in fabricated parts of various aircraft turbine engines.

The chemical composition and heat treatment of Alloy 718 act together in producing the properties desired. In particular, it has been found that the columbium, aluminum, titanium, and carbon content have important influences. Recent information indicates that for optimum creep-rupture properties, it is desirable to maintain aluminum on the high side, and titanium on the low side, at the same time using 1750 F annealing temperatures. For optimum short-time tensile properties at cryogenic or medium temperatures, the trend has been towards lower aluminum content, higher titanium content, and an annealing temperature of 1950 F. In all types of applications there has been a tendency toward reducing the range of allowable composition.

Microstructure and microconstituents are, as expected, profoundly influenced by composition and heat treatment. The report describes microstructures and conditions for the formation and solution of Laves phase (freckles), Ni_3Cb, gamma-prime strengthening phase, and other microconstituents. The nature of the γ' strengthening phase is indicated to be a metastable phase based on the Ni_3Cb composition, but with a body-centered tetragonal Ni_3V structure. After overaging this transforms to the stable orthorhombic Ni_3Cb.

Though much has been learned concerning Alloy 718, this report indicates that much more understanding needs to be obtained in order that the full potentialities of the alloy may be realized.

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INTRODUCTION

Some 5 years ago a wrought, weldable nickel-chromium-base alloy was introduced.(1)* Primarily, it was intended for use at medium temperatures, 1300 F at most, its main applications being in lightweight welded frames and other assemblies in aircraft turbojet engines. It filled a need for a weldable, wrought material which the nickel-base superalloys were not fully capable of satisfying. Though the nickel-base superalloys have outstanding strength at higher temperatures - between 1500 and 1800 F - they are quite difficult to weld.

Since Alloy 718 was first introduced, it has broadened its areas of application into the cryogenic-temperature field. Its good properties in the range of temperature from -423 to 1300 F make it especially suitable for use in the LOX-LH_2 rocket engines. In these applications, Alloy 718 is used for the fuel/oxidizer injector plates, forged rings, thrust chamber jackets, turbine wheels, bellows, and tubing.

The nominal composition of Alloy 718, contrasted with René 41, is as follows:

	Ni	Co	Cr	Fe	Mo	Ti	Al	Cb	C	B
Alloy 718	53	--	19	19	3	0.8	0.6	5.2	0.05	0.004
René 41	55	11	19	--	10	3	1.5	--	0.09	0.005

This illustrates how Alloy 718 differs from the nickel-base superalloys: (1) substitution of columbium for much of the aluminum and titanium, (2) the introduction of almost 20 percent iron, and (3) reduction of the amounts of cobalt and molybdenum.

The effect of these differences is to reduce the high-temperature capabilities of Alloy 718, but in return for this loss Alloy 718 has gained weldability.

The improved weldability of Alloy 718 derives mainly from a change in composition of the principal strengthening phase, γ' (gamma prime). While the nickel-base superalloys are strengthened by a γ' phase corresponding to $Ni_3(Al, Ti)$, the γ' in Alloy 718 is mainly $Ni_3(Al, Ti, Cb)$, or perhaps $Ni_3(Al, Ti, Cb, Mo)$. Some disagreement exists as to the exact composition of the phase. In most general terms, it is described as a metastable structure, rich in columbium, and initially precipitating such that it is coherent with the fcc (face-centered cubic) matrix. Because the rate of precipitation of the γ' is relatively low, in comparison with the rate in the nickel-base superalloys, precipitation hardening does not occur during the welding cycles. It is this fact that accounts for the good weldability of Alloy 718.

The properties and microstructure of Alloy 718 are strongly influenced by heat treatment and composition. This report discusses these influences, and to the extent that information has been found, illustrates their interrelationships. However, no attempt has been made to present a complete compilation of design properties in this report.

*References are given on pages 23 and 24.

EFFECT OF COMPOSITION AND HEAT TREATMENT ON MECHANICAL PROPERTIES(1-7)

The composition range for Alloy 718 given in Specification AMS 5596A is as follows:

Cr	17.00-21.00	Mn	0.35 maximum
Ni + Co	50.00-55.00	Si	0.35 maximum
Mo	2.80-3.30	P	0.015 maximum
Cb + Ta	5.00-5.50	S	0.015 maximum
Ti	0.65-1.15	Co	1.00 maximum
Al	0.40-0.80	Cu	0.10 maximum
B	0.0020-0.0060		
Fe	Balance		
C	0.03-0.10		

Various companies have issued specifications for Alloy 718 with chemical composition differing from that shown above. These specifications are discussed later.

Heat treatment of Alloy 718 to develop tensile properties, stress-rupture properties, or good notch-tensile properties have undergone considerable change since 1960. For background, some of the changes that have occurred are discussed below.

Characteristically, high strength in Alloy 718 is developed by a high-temperature annealing treatment followed by a lower temperature aging treatment. The specific annealing and aging temperature and times, as well as the rates of cooling from these temperatures, have been altered steadily over the past 5 years.

In 1960 the International Nickel Company,(2) who had developed the alloy, recommended that hot-rolled or annealed (mill annealed) products be aged at 1325 F for 16 hours. An optimum aging temperature of 1275 F for 16 hours was recommended for cold-rolled sheet.

Annealing temperatures of about 1750 F were recommended, and users were cautioned not to use annealing temperatures exceeding 1800 F. Figure 1 illustrates some of the data supporting the recommendations.

Subsequently, it was found that improved mechanical properties could be obtained by modifying the aging treatment.(1,3,4) Barker(1,3) reported that aging at 1200 F for 200 hours (following the 1700 F to 1325 F treatment) could raise the room-temperature tensile strength from 180,000 to 240,000 psi. Also, a double-aging procedure has been found to be beneficial. The

May, 1961, data report of the International Nickel Co.[4] recommended the following for improving the yield and tensile strengths without decreasing the ductility or stress-rupture properties:

(1) Anneal, age at 1325 F for 8 hours, furnace cool at the rate of 20°/hr to 1150 F, air cool

(2) Anneal, age at 1325 F for 8 hours, furnace cool at the rate of 100°/hr to 1150 F, hold at 1150 F for 8 hours, air cool.

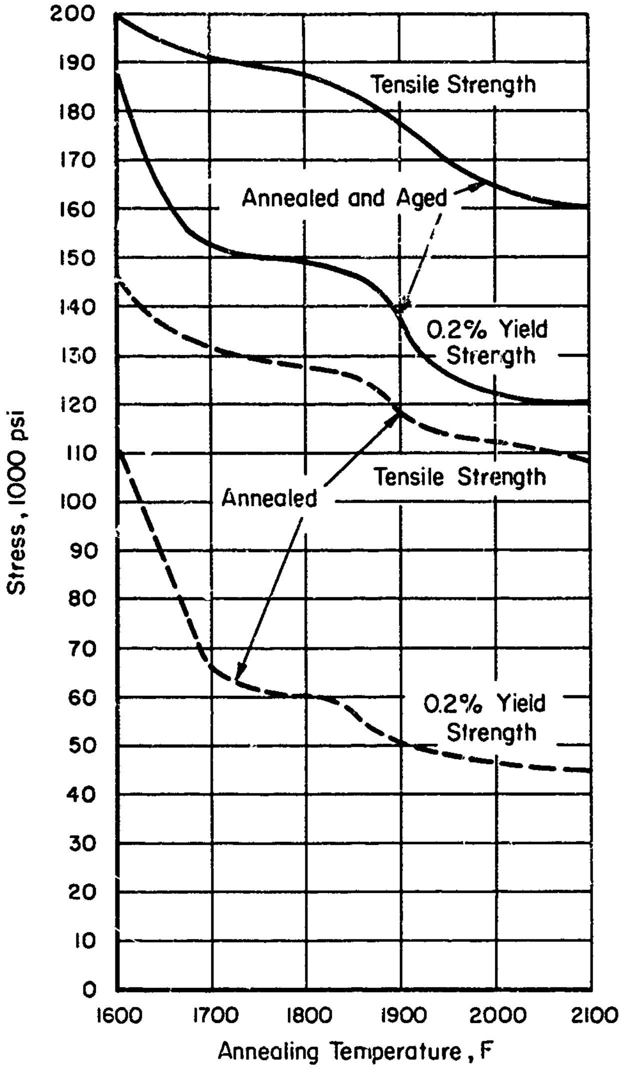


FIGURE 1. EFFECT OF ANNEALING TEMPERATURE ON THE ROOM-TEMPERATURE TENSILE PROPERTIES OF ALLOY 718 - ADAPTED FROM A 1960 BROCHURE[2]

Annealed for 15 min; aged at 1325 F for 16 hr.

In addition, a change was made in the annealing procedure. Thus, for optimum tensile properties, annealing at 1700 to 1800 F was recommended, but for best stress-rupture properties 1900 F was preferred.

In the course of the past few years, considerable data have been accumulated showing the effects of annealing temperature, aging temperature and times, and chemical composition. Nevertheless, the present situation is that the matter is still unsettled. One reason is, undoubtedly, that the data have not always been consistent. One reason for the lack of consistency seems to have been that the optimum heat treatment depends on the chemical composition, particularly the aluminum content. This is illustrated in Figure 2, which is plotted from data in a Latrobe Steel Company report.[5]

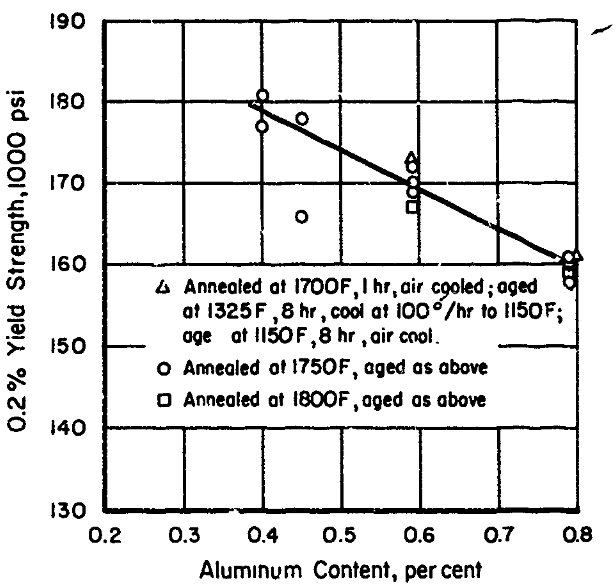


FIGURE 2. YIELD STRENGTH OF ALLOY 718 AS A FUNCTION OF ALUMINUM CONTENT

Latrobe Steel Company data[5]

Eiselstein[6] made a systematic study of the effects of changes in the titanium, aluminum, boron, and columbium content on the mechanical properties of Alloy 718. He found, as expected, that the effects of chemical composition were dependent on the heat treatment. The effect of aluminum content on the room-temperature yield strength was a function of both the annealing temperature and the aging temperature. When the alloy was heat treated as follows:

Anneal: 1750 F, 1 hr, air cool
Age : 1325 F, 8 hr, furnace cool at 20°/hr to 1150F, air cool.

it was found that increasing the titanium and columbium content within the specification range increased the yield and tensile strengths at room temperature and at 1200 F. The elongation was correspondingly decreased. Increasing aluminum, on the other hand, seems to have lowered the tensile and yield strengths, without affecting the elongation. Boron had a slight adverse effect on room-temperature tensile strength but increased the elongation.

At 1300 F, increasing the titanium content increased the tensile and yield strengths, while decreasing the elongation. Increasing the aluminum content over 0.7 percent resulted in a slight increase in tensile strength, accompanied by a decrease in the elongation.

Table 1 summarizes the results of some of Eiselstein's experiments on the effects of aluminum and titanium.

Figures 3 and 4[7] show the effect of aluminum content on the room-temperature tensile properties of Alloy 718 annealed at 1750 or 1950 F, and then aged at 1325, 1350, 1375, or 1400 F, followed by aging at 1200 F. These figures show how intimately related are the heat treatment and composition. For the highest yield strength, the optimum combination would be: low aluminum, 1950 F anneal, and a 1325 or 1350 F aging treatment. However, in rupture testing, specimens with this chemical composition and heat treatment are notch brittle.

The combination of low aluminum content, high annealing temperature, and relatively low aging temperature is also good with respect to fatigue life and short transverse properties but is not desirable when optimum long-time stress-rupture properties are needed. If the aluminum content is not low, however, the 1750 F anneal in combination with a 1325 to 1350 F aging treatment seems preferable.

The effect of columbium content on the tensile properties is shown in Figure 5. Often, when speaking of columbium, the term columbium + tantalum is used. In such a case, the tantalum content can be considered to be 10 percent of the columbium content. The figure shows a regular increase in yield strength as the columbium + tantalum is increased from 2 to 6 percent, for the annealed and aged material. As annealed (not aged), an increase in tensile and yield strengths was also observed, which is indicative of some solution strengthening.

Although Figure 5 shows some advantage in columbium contents above 5 percent with respect to tensile strength, the ductility drops noticeably. Therefore, composition limitations were first set at nominally 5 percent columbium. Existing specifications allow up to 5.50 percent columbium.

Carbon in amounts between 0.01 and 1 percent was found to reduce the yield and tensile strengths at 1300 F. Presumably, this decrease is the result of reduction in the effective amount of columbium, which is tied up by the carbon.

Smooth-bar rupture life appears to depend more on the annealing temperature than on the chemical composition. In investigating the effects of boron, titanium, aluminum, carbon, and annealing temperature (+ 1325 F age, 8 hr, furnace cooled to 1150 F, air cooled), Eiselstein[6] found that increasing the annealing temperatures from 1750 or 1800 F to 1900 F increased the rupture life more than did compositional changes while the annealing temperature was held at 1750-1800 F. With the 1900 F anneal, compositional variation was more important than when the lower annealing temperature was used. In general, it

TABLE 1. SUMMARY[a] OF THE EFFECT OF VARIATION OF THE TITANIUM AND ALUMINUM CONTENT OF ALLOY 718[6]

Element	Range of Variation, percent	Tensile Strength Room Temp	Tensile Strength 1200 F	Tensile Strength 1300 F	Yield Strength Room Temp	Yield Strength 1200 F	Yield Strength 1300 F	Elongation Room Temp	Elongation 1200 F	Elongation 1300 F
Titanium	0.6-1.3	+	+	+	+	+	+	-		-
Aluminum	0.4-0.9	-	-		-	-		0		
	0.1-0.8			+ (Over 0.7%)		~0				-

(a) +, increase; -, decrease; 0, no change.
Heat treatment: 1750 F, 1 hr, AC + 1325 F, 8 hr, FC at 20°/hr to 1150 F, AC.

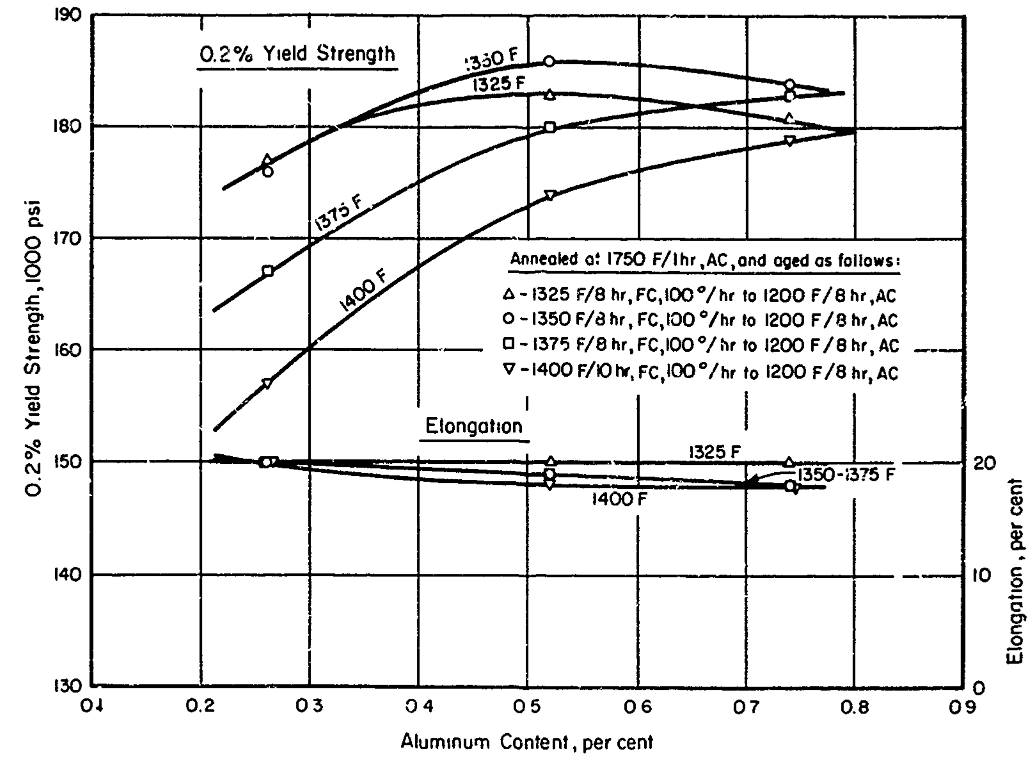


FIGURE 3. EFFECT OF ALUMINUM CONTENT ON THE ROOM-TEMPERATURE YIELD STRENGTH OF ALLOY 718 HOT-ROLLED BAR STOCK

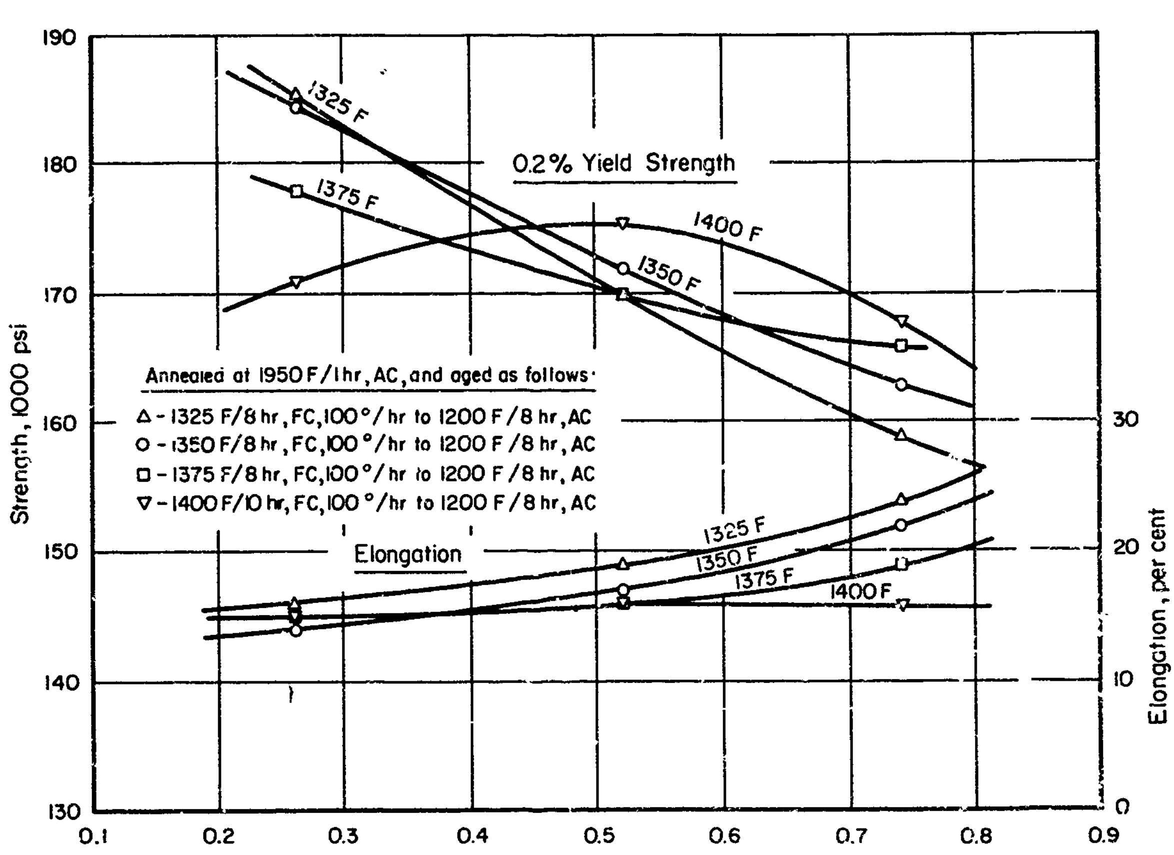


FIGURE 4. EFFECT OF ALUMINUM CONTENT ON THE ROOM-TEMPERATURE 0.2% YIELD STRENGTH OF ALLOY 718, HOT-ROLLED BAR STOCK

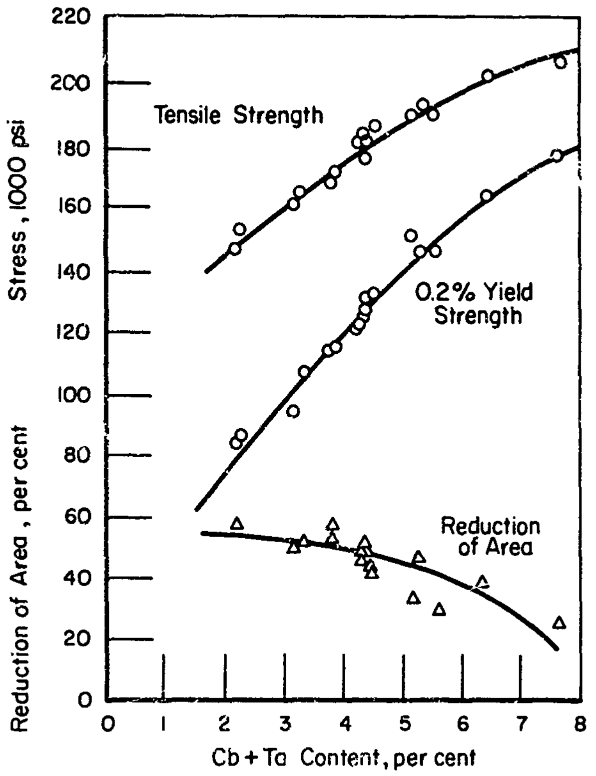


FIGURE 5. ROOM-TEMPERATURE TENSILE PROPERTIES OF ALLOY 718 AS A FUNCTION OF COLUMBIUM AND TANTALUM CONTENT (After Eiselstein)(6)

was found that in 1300 F, 75,000-psi rupture tests

(1) Titanium had little or no effect on the smooth-bar rupture life.

(2) Increased aluminum increased the rupture life, but there was no advantage to adding over 0.4 percent.

(3) Boron increased the rupture life, but for annealing temperatures of 1750 or 1800 F, there was no advantage to adding over 0.003 percent. For annealing temperatures of 1900 F, however, amounts of 0.004-0.006 were most effective, particularly when the aluminum was low and the titanium high.

(4) Carbon, in the presence of 0.004-0.006 percent boron, was found to increase the notch-bar rupture life but decrease the smooth-bar rupture life.

Data on the interrelationships among chemical composition, heat treatment, and mechanical properties are still being gathered. The foregoing gives some idea of the complexity of the relationships, and suggests why more data are needed. Meanwhile, specifications for chemical composition and heat treatment have undergone many changes since the alloy was first developed. These specifications are discussed in the next section.

TRENDS IN SPECIFICATIONS

Specifications for Alloy 718 have been issued by a number of companies,(8-15) who intend using the alloy in different kinds of service. In the main, the service for which the alloy is intended can be grouped into the following categories:

(1) High temperature, requiring good short-time tensile properties

(2) High temperature, requiring good creep-rupture properties

(3) Cryogenic, requiring good tensile properties and toughness.

The major application for the material in the first category is in the hot parts of liquid-fueled rocket engines. Aircraft turbine engines are the main applications for the second category. In Category (3) fall such missile hardware as cryogenic tankage and piping. These applications are sometimes called out in the specifications themselves, as indicated in the following quotations from some of the specifications:

AMS 5596A(8)

"2. Application: Primarily for parts, such as cases and ducts, requiring high resistance to creep and stress rupture up to 1300 F (705 C) and oxidation resistance up to 1800 F (980 C), particularly those which are formed and then heat treated to develop required properties."

RBD170-101(10)

"1. Scope: This material is a nickel-base heat-resistant alloy which is intended primarily for parts requiring high short-time tensile strength up to 1000 F and oxidation resistance up to 1800 F. It has good cryogenic properties and better weldability than other age hardenable nickel-base alloys."

EMS-581c(15)

"3. Application: Primarily for parts requiring high strength and corrosion resistance at both cryogenic and elevated temperature, particularly those which are machined and welded and then heat treated to develop required properties. Material has good oxidation resistance up to 1800 F but is useful at temperatures above 1200 F only when stresses are low."

Because of the necessity of satisfying different kinds of applications, we find today that the specifications are becoming more restrictive as to chemical composition and, depending on the intended application, will ask for widely differing heat treatments.

In particular, the differences between the various specifications lie mainly in the columbium, aluminum, and titanium content. Smaller differences exist among the carbon and boron content. Table 2 illustrates the allowable chemical composition according to selected specifications.

The general trend seen in Table 2 is for the specifications applicable to aircraft engines [8,12-14] to tend toward:

(1) Limits of 0.65-1.15 titanium

(2) Limits of 0.40-0.80 aluminum

(3) Limits of 5.00-5.50 columbium.

The trend for liquid-fuel rocket engine manufacturers[9,10] is towards higher titanium (up to 1.4 percent) and lower aluminum content (0.2 to 0.7 percent).

A good illustration of the developments that have taken place in arriving at the present specifications is obtained by comparing Rocketdyne Specification RB0170-039 (November, 1962) with RB0170-101 (March, 1965) which replaces -039. Notice that the columbium range has been decreased by raising the minimum from 4.75 to 5.00 percent; the titanium range has been decreased by raising the minimum from 0.65 to 0.85 percent; the aluminum range has been decreased by raising the minimum from 0.35 to 0.40 and lowering the maximum from 0.85 to 0.70. Also, the maxima on carbon, silicon, and manganese have been lowered.

A further example of this trend is observed by comparing AiResearch Specification EMS-581c (January, 1965) with its original version, EMS-581 (August, 1962). The main application is for cryogenic tankage, but may include high-temperature tensile-limited components.

	EMS-581	EMS-581c
Cb + Ta	4.75-5.75	4.75-5.50
Ti	0.30-1.30	0.70-1.40
Al	0.20-1.00	0.20-0.70
B	--	0.006 max
C	0.10 max	0.08 max

Besides changes in composition, there have been changes in heat treatment. Most curious has been the complete reversal in the original ideas

TABLE 2. CHEMICAL COMPOSITION OF ALLOY 718 ACCORDING TO VARIOUS SPECIFICATIONS

Specification Identification	Company	Amount Specified[a], percent								
		Cb + Ta	Ti	Al	B	C	Si (max)	Mn (max)	S (max)	Cu (max)
AMS 5596[8][b]	Society of Automotive Engineers	5.00-5.50	0.65-1.15	0.40-0.80	0.002-0.006	0.03-0.10	0.35	0.35	0.015	0.10
AGC-44152[9]	Aerojet-General	4.75-5.5	0.65-1.40	0.10-0.80	0.001-0.010	0.10 max	0.45	0.45	0.015	0.30
RB0170-101[10]	North American Aviation-Rocketdyne	5.00-5.50	0.85-1.15	0.40-0.70	0.006 max	0.06 max	0.35	0.35	0.015	0.30
RB0170-039[11]	North American Aviation-Rocketdyne	4.75-5.50	0.65-1.15	0.35-0.85	0.006 max	0.03-0.10	0.45	0.40	0.015	0.15
B50T69-S6[12]	General Electric Company, Large Jet Engine Division	4.75-5.50	0.70-1.40	0.20-0.80	0.002-0.010	0.10 max	0.45	0.35	0.03	0.75
C50T79(S1)[13]	General Electric Company, Large Jet Engine Division	5.00-5.50	0.65-1.15	0.40-0.80	0.002-0.010	0.10 max	0.40	0.35	0.03	--
PWA 1009-C[14]	Pratt and Whitney Aircraft	5.00-5.50	0.65-1.15	0.40-0.80	0.006 max	0.03-0.10	0.35	0.35	0.015	0.10
EMS-581c[15]	AiResearch Manufacturing Company	4.75-5.50	0.7-1.4	0.2-0.7	0.006 max	0.08 max	0.45	0.40	0.015	0.30

(a) In addition to the elements shown in the table, all specifications call for the following:
Co, 1.00 max, Ni + Co, 50.00-55.00, Cr, 17.00-21.00, Mo, 2.80-3.30, Fe, balance.
When specified, P is 0.015 maximum. Ta is listed in RB0170-101 as 0.50 max and in B50T69-S6 as 1.00 max.

(b) Superscripts refer to references.

that high annealing temperatures were good for creep-limited applications, and low annealing temperatures for tensile-limited applications. The aircraft engine manufacturers, desiring good creep-rupture life, have found that 1700 or 1750 F for 1 hour is the preferred annealing treatment.* On the other hand, when good tensile properties at temperatures to about 1200 F are needed, the annealing temperature is now specified as 1950 F. This temperature seems preferred also when toughness at cryogenic temperatures will be required in service.

The main reason for not using the 1950 F anneal for creep-rupture-limited applications is that the material lacks rupture ductility. The trend toward higher annealing temperature for tensile-limited applications has been tied in with a lowering of the specified aluminum content. Data illustrating the improved tensile properties under these conditions were shown in the previous section. Moreover, it has been reported that lowering the aluminum content also improves the weldability.

Typical heat treatments are listed in Table 3. Generally, the heat treatment consists of an anneal for 1 hour or more, followed by air cooling (or faster*). Then the alloy is double aged to develop high strength (see page 22 for an explanation of the mechanism). The usual method is to hold for 8 hours at the first aging temperature, furnace cool at the rate of 100°/hr to the second aging temperature, hold 8 hours, and air cool. As an alternative, some specifications permit holding 8 hours at the first aging temperature, furnace cooling at an unspecified rate, and holding at the second aging temperature until the total elapsed time since the start of the first aging step is 18 hours. In the Rocketdyne and Aerojet-General specification, the time of first aging may be 10 hours, and the total elapsed time may be 20 hours instead of 18 hours.

*Sometimes the annealing treatment is referred to as a "solution treatment." This is proper only when the temperatures exceed 1750 F, because complete solution does not take place below 1750 F (see next section).

Depending on section size, the specifications call for the following mechanical properties after the aging treatment:

	Minimum Yield Strength, psi
At room temperature	145-150,000 psi
At 1200 F	120-125,000 psi
At -320 F	175,000 psi
Rupture stress for 23-hour life at 1300 F	72,500-75,000 psi

*Water quenching, oil quenching, or air cooling (~400°/min). Slow cooling (<40°/min) can result in low yield strengths after aging.(16)

TABLE 3. ANNEALING AND AGING TEMPERATURES

Specification Identification	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 Company	1700	1325	1150	I
C50T79(S1)	General Electric Company	1800	1325	1150	I
PWA 1009-C	Pratt and Whitney Aircraft	1750	1325	1150	I or II
EMS-581c	AiResearch	1950	1350(b)	1200	I
RB0170-101	Rocketdyne	1950	1400	1200	III
AGC-44152	Aerojet-General	1950	1350	1200	IV

(a) I: Hold 8 hr at first aging temperature, furnace cool at 100°/hr to second aging temperature. Hold 8 hr, air cool.
II: Hold 8 hr at first aging temperature, furnace cool to second aging temperatur 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.

There are slight variations from these figures in some of the specifications, but they illustrate the general requirements. When the 1950 F anneal is specified, however, no demands for a 1300 F rupture test are made.

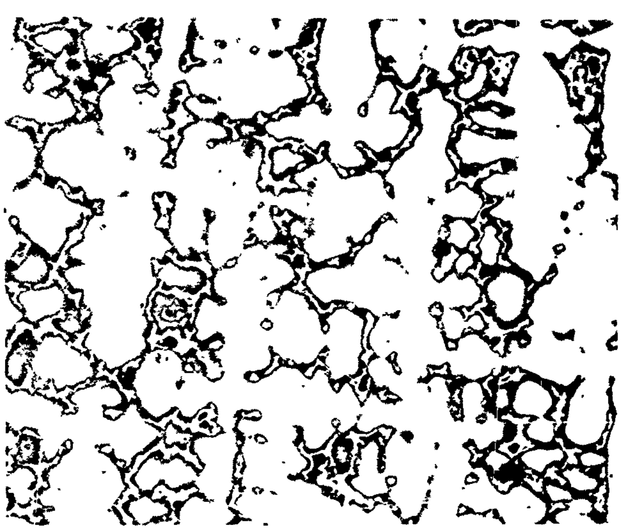
The effect of various heat treatments on properties has been accompanied by a considerable amount of work on understanding the microstructure and phases that are present in Alloy 718. These are discussed in the next section.

MICROSTRUCTURE AND MICROCONSTITUENTS

Alloy 718 is used primarily in the wrought form. Nevertheless, certain features of the cast structure can be retained in the wrought structures, and, in these cases, have a strong influence on the mechanical properties of the wrought product. Accordingly, the microstructures of both cast and wrought Alloy 718 are discussed.

Cast Alloy

Figures 6 and 7 show typical cast structures obtained from different parts of the same ingot.[6] Figure 6 shows the head end, while Figure 7 shows the toe end, which had cooled much faster than the head end. The dendrites are the light areas and the matrix is the dark area. The light constituent within the dark area has been identified as a Laves phase of the A_2B type. Figure 8 shows a finer distribution of the phase as found in the center of a 9-inch-diameter air-melted ingot.

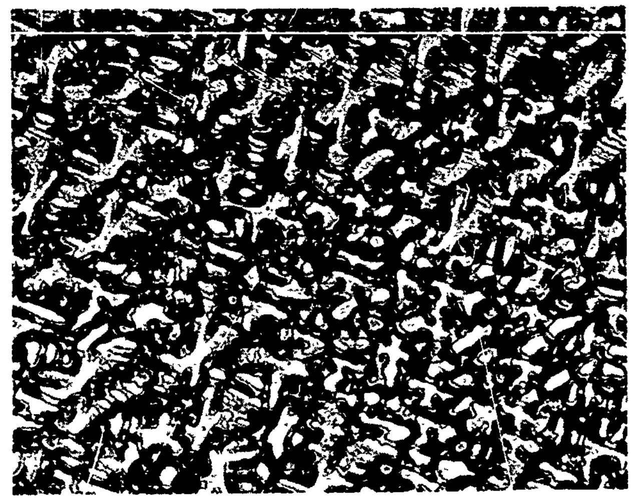


100X

Overetched in Chromic Acid

FIGURE 6. CAST STRUCTURE OF HEAD OF INGOT WITH 5.4% Cb + Ta[6]

(Reduced approximately 20 percent in printing)

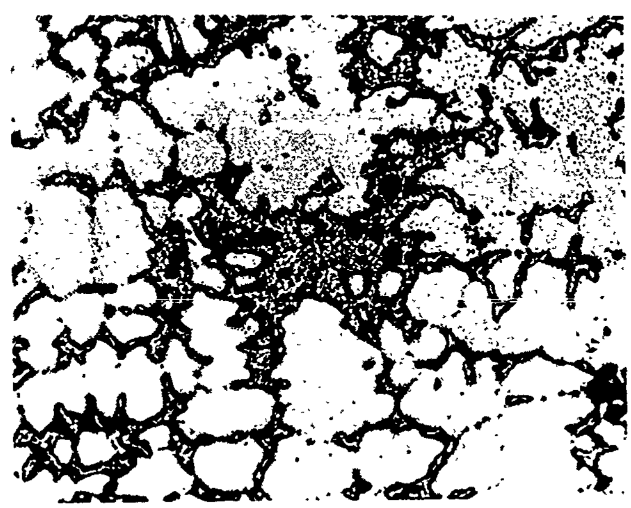


100X

Overetched in Chromic Acid

FIGURE 7. CAST STRUCTURE OF TOE OF INGOT WITH 5.4% Cb + Ta[6]

(Reduced approximately 20 percent in printing)



100X

Electrolytic, Chromic Acid

FIGURE 8. CENTER OF AIR-MELTED 9-INCH-DIAMETER INGOT[22]

(Reduced approximately 20 percent in printing)

Freckles

The Laves phase has been identified with the phenomenon of "freckles" which (unlike the microstructures shown) appear as dark-etching constituents when large sections are macroetched. Figure 9[6] is an example of freckles in a large forging in which the cast structure has been retained in the center. Their geometry is actually rod-like, extending in a general axial direction but perpendicular to the freezing front in arc-cast billets. Eiselstein[6] reported that in the annealed

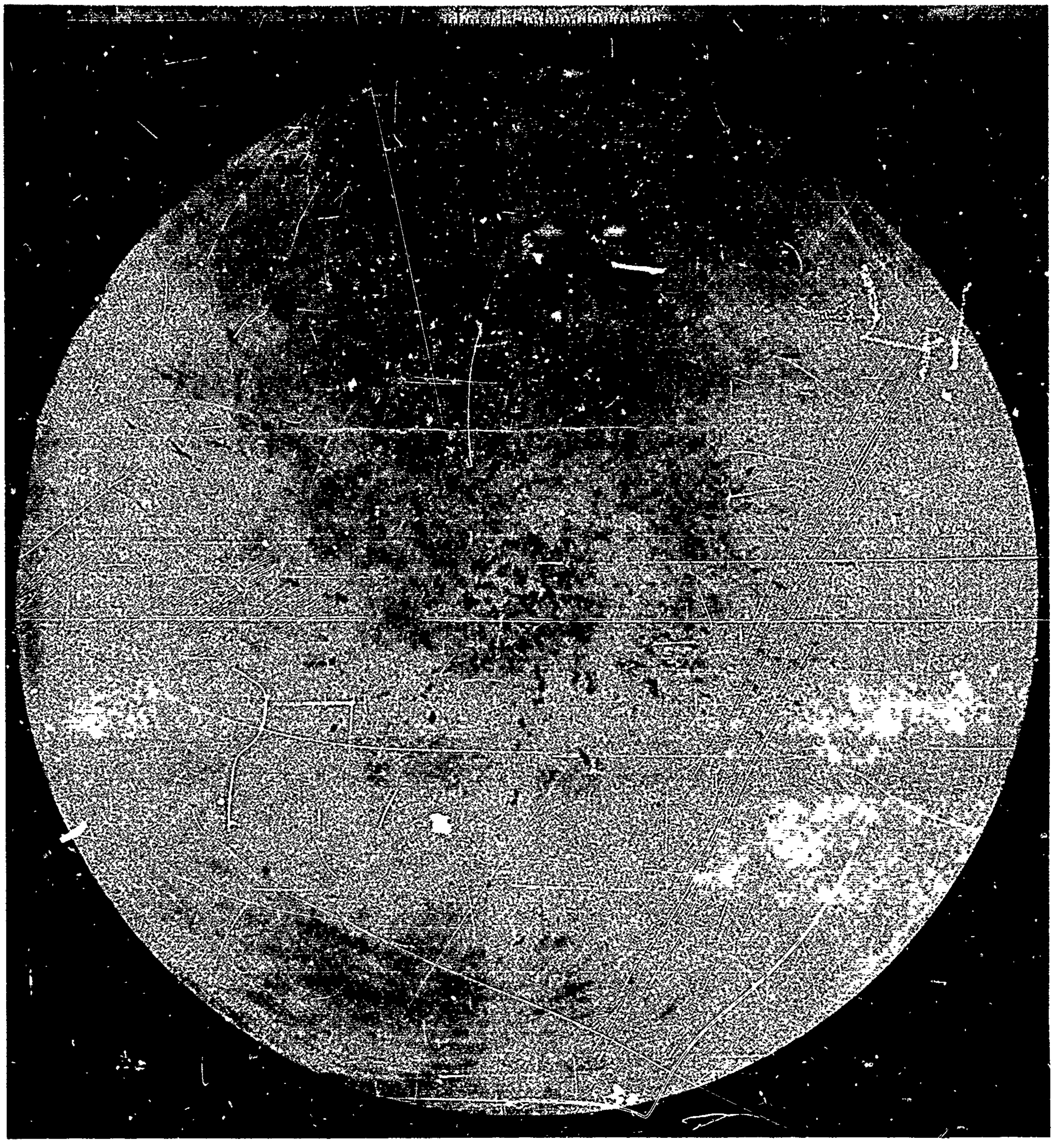


FIGURE 9. VACUUM-ARC-MELTED AND CAST 16-IN.-DIAMETER INGOT FORGED AND TURNED TO 12-IN.-DIAMETER ROUND

This cross section was taken from the toe end of the forged ingot. Note the dark-etching "freckles".(6)

and aged condition, samples containing freckles had yield strengths of approximately 120,000 psi, as compared to an expected value of 150,000. It was also reported that the ductility of the freckled samples was "poor".

Poor ductility in regions having a large amount of Laves phase in a coarse dendritic structure seems well documented. Figures 10 and 11 are examples. The specimens represented in Figures 10 and 11 were from slightly worked cast ingots. The elongations in tensile tests were, respectively, zero percent and 15 percent.(6)

Figures 12 and 13 show areas in a forging made by Beech Aircraft Corporation(17) from a 12-in. forged billet of Alloy 718 that had shown severely freckled areas. The banded structure and mixed grain size is believed to have resulted from the chemical heterogeneity in the vicinity of the freckles.

Because of the adverse effect of freckles (or Laves phase), there has been a good deal of study put into identifying the conditions of chemical composition and heat treatment that promote them. Generally, the occurrence of the Laves phase has been associated with the cast state.

Barker(18) has reported on the effect of various heat treatments of the microstructures and microconstituents of cast Alloy 718. Figures 14-20 show some of the microstructures obtained from various cast ingots. These microstructures show that the Laves phase present in the cast structure is not affected by solution treatment at temperatures below 2100 F. Apparently it can be dissolved at 2100 F or above. Kaufman and Palty(19) obtained an $Fe_2(Ti, Cb)$ phase in wrought Alloy 718 annealed at 2250 F. It is likely, however, that grain-boundary melting had occurred at this temperature, as shown in Figure 21. Subsequent treatment at 1800 and 2000 F tended to spheroidize and agglomerate the phase. Eiselstein(6) reported that this phase was present in a specimen held for 100 hours at 1700 F, indicating that the Laves phase will appear after long-time exposure to relatively high temperatures.

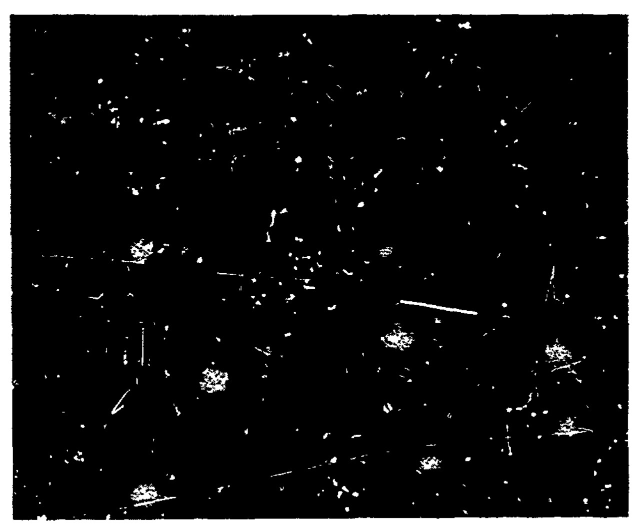
The Laves phase associated with the phenomenon of freckles has been found to be isomorphous with Fe_2Ti. It has been described as $Fe_2(Ti, Cb)$(19) or simply $M_2(Ti, Cb)$.(18) Eiselstein(6) found that the residue obtained by dissolving a freckled region in a forging was high in columbium, molybdenum, and nickel and contained, in addition, small amounts of iron, chromium, and titanium (12 to 15 weight percent in all). The nickel content increased when the specimen was annealed and aged.

Barker(18) also presented data of other investigators which showed the residues in annealed and aged specimens to be high in nickel. It would seem that the actual chemical composition of the Laves phase is subject to considerable variation and depends strongly on the thermal history of the specimen. Certain of the data presented in Barker's report indicate an approximate formula of the following form:

100X Overetched in Chromic Acid

FIGURE 10. SLIGHTLY WORKED CAST STRUCTURE HAVING ZERO PERCENT ELONGATION IN A TENSILE TEST(6)

(Reduced approximately 20 percent in printing)



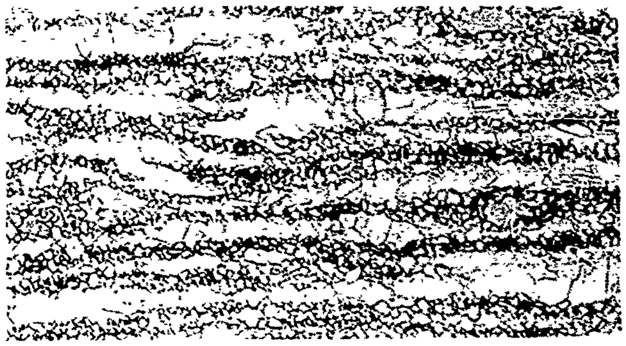
100X Overetched in Chromic Acid

FIGURE 11. SLIGHTLY WORKED CAST STRUCTURE HAVING 15 PERCENT ELONGATION IN A TENSILE TEST(6)

(Reduced approximately 20 percent in printing)

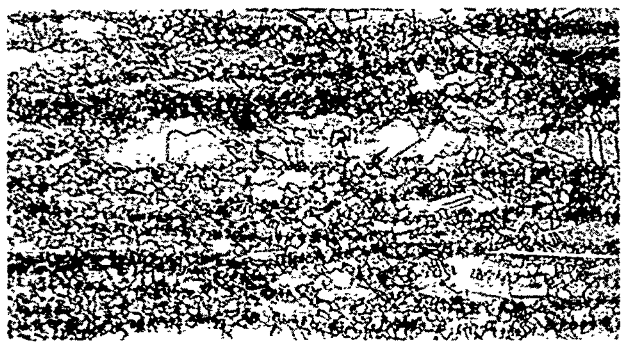
100X

a.



100X

b.



100X

c.

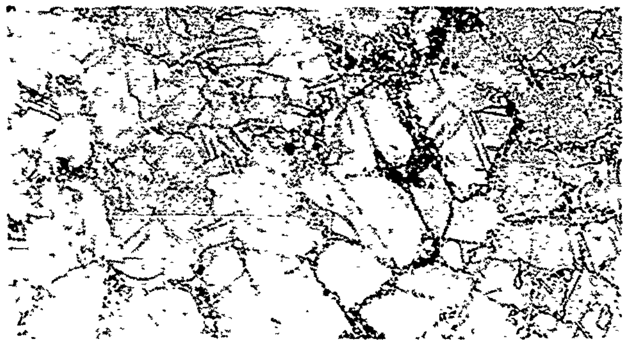
FIGURE 12. MICROSTRUCTURE IN THE BOSS AREA OF A HEMISPHERICAL TANK FORGING(17)

Etchant: mixed acids.

(Reduced approximately 20 percent in printing)

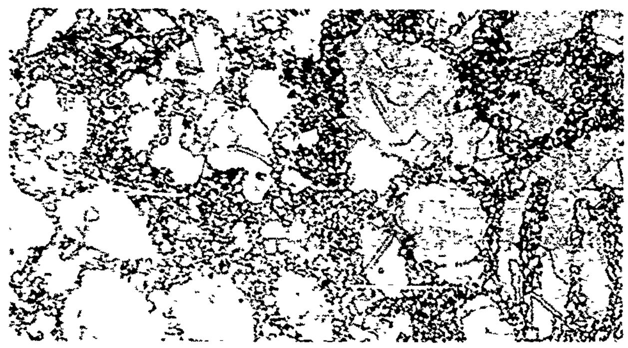
100X

a.



100X

b.



100X

c.

FIGURE 13. MICROSTRUCTURE IN THE BODY OF A HEMISPHERICAL TANK FORGING(17)

Etchant: mixed acids.

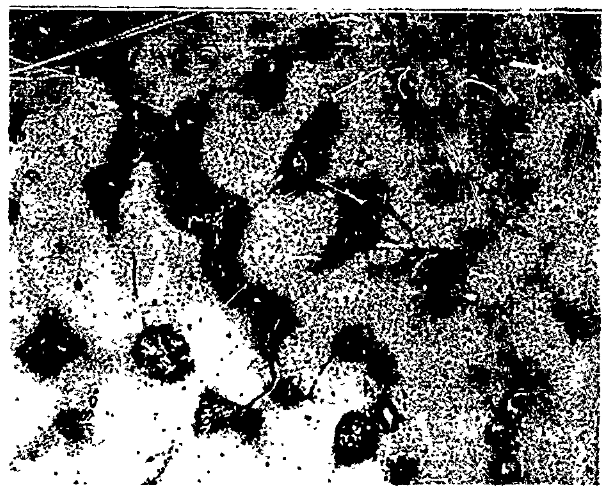
(Reduced approximately 20 percent in printing)

250X

FIGURE 14. AS-CAST ALLOY 718

The irregular white phase is Laves. The blocky phases are carbides and nitrides. The dark etching phase is believed to be Ni_3Cb(18,19)

(Reduced approximately 20 percent in printing.)

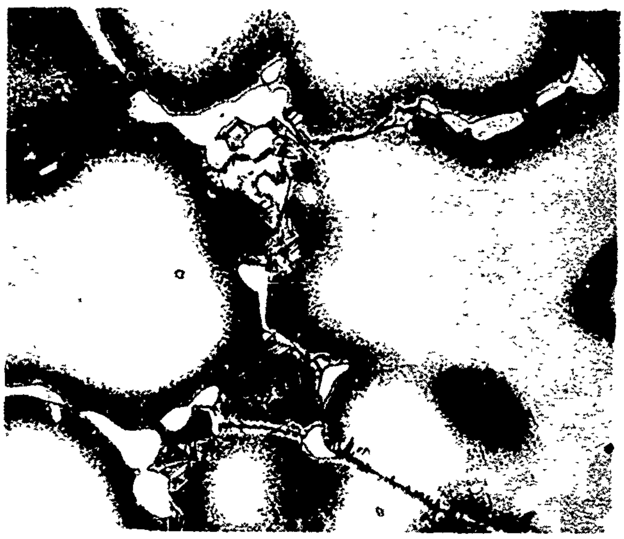


250X

FIGURE 15. CAST ALLOY 718 ANNEALED AT 1700 F, 1 HR, AIR COOLED, THEN AGED AT 1325 F FOR 16 HR

Some of the Ni_3Cb has become acicular.(18,19)

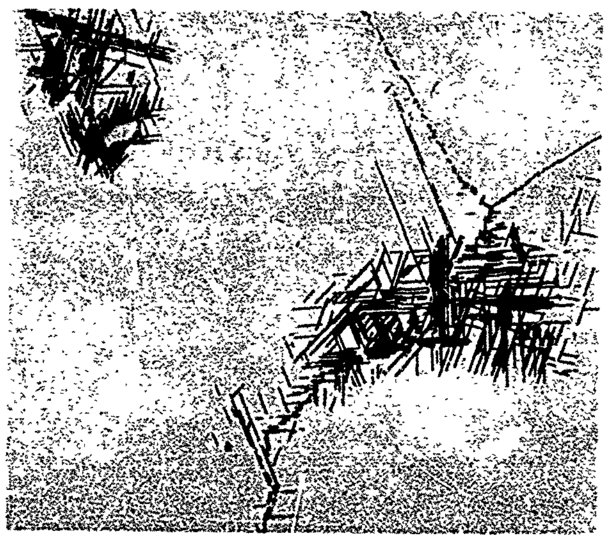
(Reduced approximately 20 percent in printing.)



500X

FIGURE 16. CAST ALLOY 718 ANNEALED AT 1700 F FOR 1 HR AND AIR COOLED(18)

(Reduced approximately 20 percent in printing.)

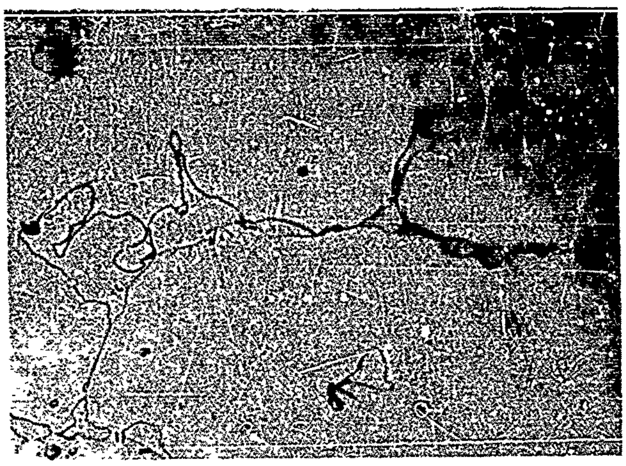


500X

FIGURE 17. CAST ALLOY 718 ANNEALED AT 1800 F FOR 1 HR AND AIR COOLED

The acicular phase is Ni_3Cb, which surrounds the Laves phase.(18)

(Reduced approximately 20 percent in printing.)

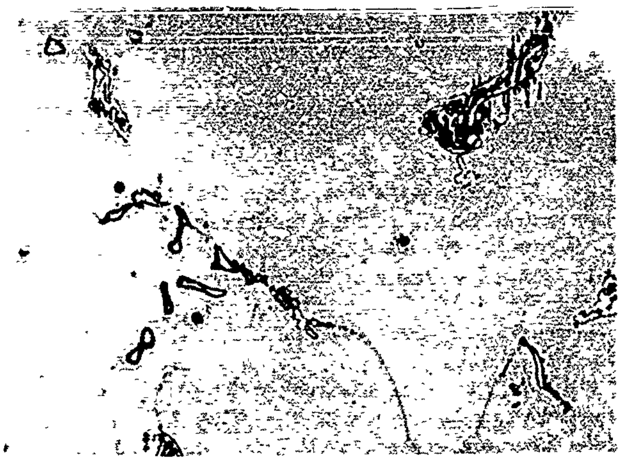


500X

FIGURE 18. CAST ALLOY 718 ANNEALED AT 2000 F FOR 1 HR AND AIR COOLED

The blocky white phase is Laves. The Ni_3Cb has been almost completely dissolved.[18]

(Reduced approximately 20 percent in printing)



500X

FIGURE 19. CAST ALLOY 718 ANNEALED AT 2000 F FOR 1 HR AND AIR COOLED

The irregularly shaped white areas are Laves phase, but the other blocky white particles are carbides of columbium and titanium.[18]

(Reduced approximately 20 percent in printing)

500X

FIGURE 20. CAST ALLOY 718 ANNEALED AT 2100 F FOR 1 HR, AIR COOLED, AND THEN AGED AT 1325 F FOR 16 HR AND AIR COOLED

Most of the white particles are Laves phase partially dissolved.[18]

(Reduced approximately 20 percent in printing)

250X

FIGURE 21. ALLOY 718 SHEET ANNEALED AT 2250 F, 2 HR, ICE-BRINE QUENCHED

Aged at 1400 F, 100 hr, water quenched.[19]

(Reduced approximately 20 percent in printing)

$(Ni_{0.6}Fe_{0.2}Cr_{0.2})_2(Cb_{0.7}Mo_{0.3})$. Other data, including Eiselstein's, fail to conform to a simple A_2B formula in which certain elements enter the A portion of the formula and other elements the B portion.

It seems apparent, also, that considerably more work should be done to characterize the Laves phase, the conditions under which it will form, and the effects of thermal treatment on it.

The temperatures and compositions at which the Laves phase will form have been depicted in Figure 22 by Eiselstein[6] in the form of a pseudobinary diagram with the Alloy 718 base and the columbium content as the two components. The diagram shows the areas in which Laves phase formed after holding specimens of the alloy at the indicated temperatures for 100 hours. It suggests that high columbium contents promote the formation of Laves phases. The actual conditions of cooling in a large ingot can result in the formation of Laves phase in local columbium-rich regions, even though the average composition may be low in columbium content. Hence, Laves phase is found in heavily cored castings in the interdendritic regions.

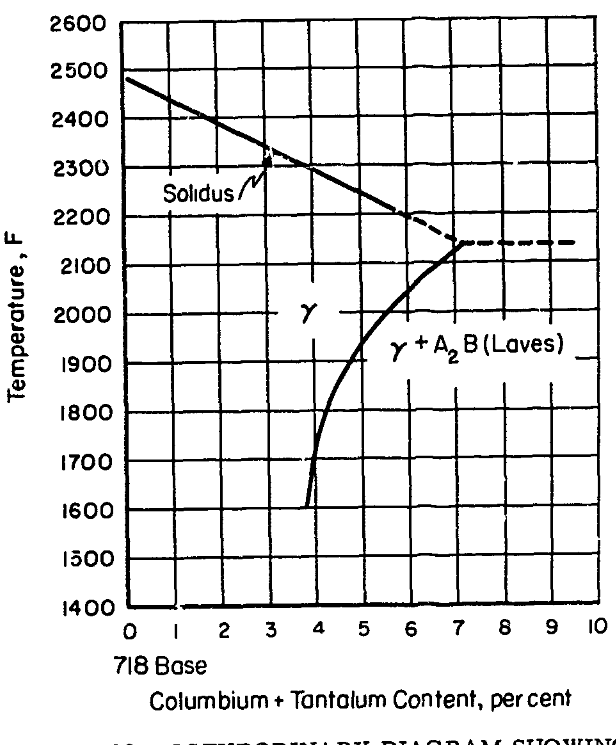


FIGURE 22. PSEUDOBINARY DIAGRAM SHOWING THE CONDITIONS FOR FORMING LAVES PHASE IN ALLOY 718 AFTER 100-HR RESIDENCE AT THE INDICATED TEMPERATURE (After Eiselstein)[6]

Other Phases in Cast Alloy 718

Kaufman and Palty[19] reported some Ni_3Cb, and carbides and nitrides of titanium and columbium in the as-cast structures. The amount of Ni_3Cb increased after the material was annealed at 1700 F (1 hour, air cooled) and aged at 1325 F (16 hours, air cooled). Figures 14-20, from Barker's report.[18] show these phases in the microstructure. In addition, it should be borne in mind that the major strengthening phase, γ' is not visible in the optical microscope. The orthorhombic Ni_3Cb, apparently, is the equilibrium phase obtained after long-time aging of the Ni_3(Cb, Mo, Ti), gamma prime.

Wrought Alloy

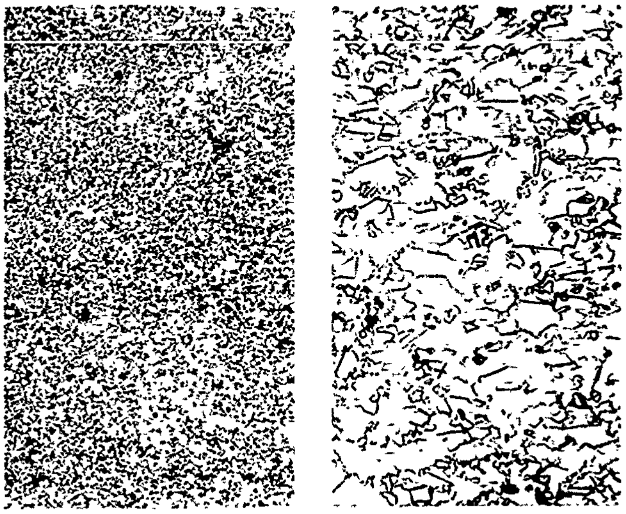
Alloy 718 bar has a microstructure typical of wrought nickel-base alloys. Figures 23-26[20] are illustrative of the microstructures found in consumable-electrode vacuum-melted stock, and show the changes produced by the annealing and aging treatments.

Aging of the annealed wrought structure at temperatures in the neighborhood of 1300 to 1400 F precipitates the γ' corresponding to Ni_3(Cb, Mo, Ti) or Ni_3(Cb, Mo, Al, Ti). The lattice parameter of the precipitated phase is about 0.8 percent larger than the lattice parameter of the fcc matrix. The resulting coherency strains account for most of the strengthening of the alloy. Aging for long times or at higher temperatures transforms the metastable γ' to the orthorhombic Ni_3Cb, which is stable.[1,6,18,19,21] This is discussed further on page 22.

Overaging for, say, 30 hours at 1400 F results in precipitation that produces appreciable darkening of the grains and some precipitation at the grain boundaries. This darkened microstructure is probably indicative of the Ni_3Cb precipitation.

Newcomer[21] examined the ability of various annealing treatments to remove the effects of overaging at 1400 F. He exposed specimens that had been overaged at 1400 F to 15-minute annealing treatments at temperatures from 1500 to 2150 F. The following paragraphs describe the microstructures observed.

The material was received in the mill-annealed condition. The microstructure showed equiaxed grains with extensive twinning characteristic of an austenitic matrix. Primary (Cb,Ti)C and nitrides were dispersed randomly throughout the grain. In addition, at high magnification (2000X), networks of spherical particles were visible that seem to outline grain boundaries from a prior structure.

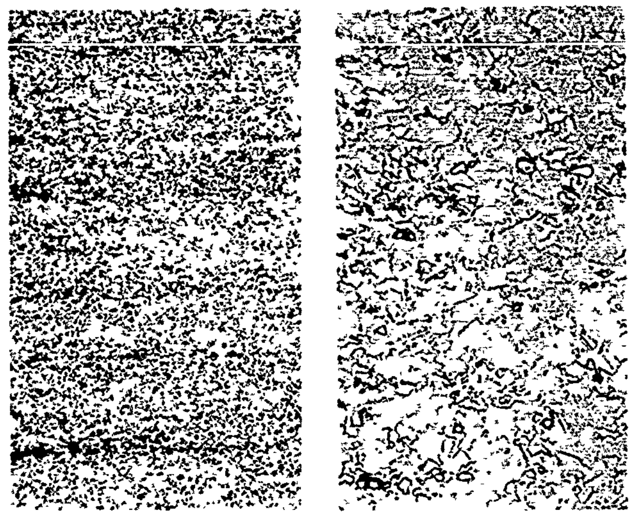


100X a. 500X b.

FIGURE 23. TRANSVERSE CROSS SECTION OF ALLOY 718 BAR

Etched with $92HCl:5H_2SO_4:3HNO_3$

(Reduced approximately 20 percent in printing)

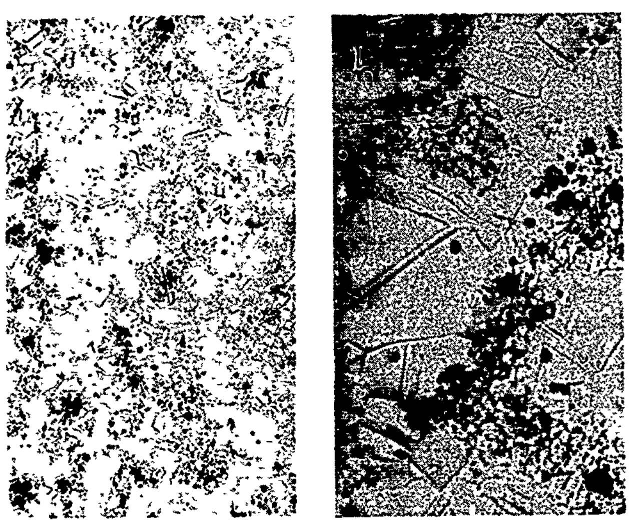


100X a. 500X b.

FIGURE 24. LONGITUDINAL CROSS SECTION OF ALLOY 718 BAR

Etched with $92HCl:5H_2SO_4:3HNO_3$.

(Reduced approximately 20 percent in printing)

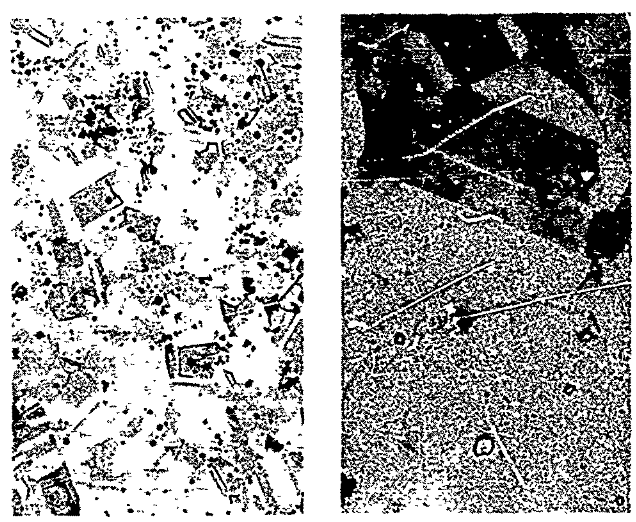


100X a. 500X b.

FIGURE 25. TRANSVERSE CROSS SECTION OF ALLOY 718 BAR ANNEALED AT 1800 F AND DOUBLE AGED

Etched with $92\ HCl:5H_2SO_4:3HNO_3$. The usual double-aging treatments are (1) 1325 F, 8 hr, furnace cool at a rate of 20°/hr to 1150 F, air cool or (2) 1325 F, 8 hr, furnace cool at a rate of 100°/hr to 1150 F. Hold at 1150 F for 8 hr, air cool. The two schedules give similar results.

(Reduced approximately 20 percent in printing)



100X a. 500X b.

FIGURE 26. TRANSVERSE CROSS SECTION OF ALLOY 718 BAR ANNEALED AT 1950 AND DOUBLE AGED

Etched with $92\ HCl:5H_2SO_4:3HNO_3$. The usual double-aging treatments are (1) 1325 F, 8 hr, furnace cool at a rate of 20°/hr to 1150 F, air cool or (2) 1325 F, to 8 hr, furnace cool at a rate of 100°/hr to 1150 F. Hold at 1150 F for 8 hr, air cool. The two schedules give similar results.

(Reduced approximately 20 percent in printing)

This structure was aged at 1400 F for 30 hours and then annealed at 1500 F, which removed some of the darkening from the grain and agglomerated some of the grain-boundary phase. Annealing at 1600 F removed most of the darkening effect and produced some extremely coarse agglomeration of Ni_3Cb at the grain boundaries. Annealing at 1700 and 1800 F produced microstructures similar to that observed in the mill-annealed material. An appreciable amount of coarse Ni_3Cb was visible at the grain boundaries, but the size and amount was less after the 1800 F treatment than after 1700 F. The coarse grain-boundary precipitate was eliminated by the 1900 F anneal, though networks of small particles, the (Cb,Ti)C, and some TiN were still visible. After the 2000 and 2100 F anneals all phases except the (Cb,Ti)C and the TiN had disappeared. Also, considerable grain growth had occurred. (Subsequent aging precipitated a phase at the grain boundaries, which could be the same as the networks observed in the mill-annealed material.) Annealing at 2150 F seems to have dissolved some of the (Cb,Ti)C and during aging reprecipitated it as films along the grain boundaries or twin planes.

The effect of annealing temperature on the grain size was found to be as follows:

Annealing Temperature, F	Average Grain Size, mm
1900 and below	0.025-0.035
2000	0.120
2100	0.150
2150	0.150-0.200

Aging of annealed wrought Alloy 718 has been studied by Kaufman and Palty,[19] Eiselstein,[6,22] and Barker.[1,18] In Kaufman and Palty's study, sheet material was annealed for 2 hours at 2250 F, ice-brine quenched and aged as follows:

Temperature, F	Time, hours
1300	100
1400	100
1500	100
1700	48
1850	24
2000	6

As mentioned in connection with Figure 21, it seems that grain-boundary melting has occurred in the 2250 F annealing treatment. This might have influenced some of the microstructures and phases found in the aged structures. They reported that the major strengthening phase is orthorhombic Ni_3Cb, though it is recognized now that this is the overaged, stable phase. In addition, $Fe_2(Ti,Cb)$ was found, as discussed previously in connection with the cast material. Angular dispersed particles of (Ti,Cb)(C,N) were found. These were unaffected by temperatures up to 2250 F. Barker,[18] citing work by Eiselstein and Radavich, suggests that the carbides and nitrides are separate, and that one phase consists of CbC with some titanium substituted for the columbium, and the other phase is TiN. Kaufman also identified a Cr_7C_3 carbide phase, but indicated that it might be associated with the $Fe_2(Ti,Cb)$ type of precipitate. It dissolved at temperatures between 1500 and 1700 F.

Figures 27a-27j show the microstructures seen at 1500X in annealed hot-rolled rod aged at various temperatures.[22] The specimens were annealed at 2200 F for 1 hour and cooled to the aging temperature, held 4 hours and water quenched.

At the 1500X magnification of the micrographs in Figure 27 no γ' has been resolved. Figure 27g, which shows the structure obtained after the 1400 F treatment, shows a great deal of precipitate. Accordingly, it is instructive to view a similar structure at higher magnifications. Figure 28 is an electron micrograph of such a structure at 10,000X. The γ' has been resolved and seems to be in the form of platelets. Many of the platelets have been cut on edge, but many can be seen as round dots. Close examination shows them to be oriented along definite planes. Figure 29 shows a somewhat coarser γ' and at higher magnification. The orientation pattern is quite evident in this specimen.

In the grain boundaries shown in Figure 28, the (Cb,Ti)C is seen. This is a grain-boundary film. In Figure 29, the film had adhered to the replica, and its filmy nature has been emphasized.

Figure 30 shows an electron micrograph of the structure obtained after the 1600 F treatment.[18] A few carbides remain in the grain boundaries, but the matrix precipitates have been dissolved.

The effects of long-time aging on the microstructure are illustrated in Figures 31-33.[6] The specimens shown here were solution treated at 2100 F (rather than the 2200 F treatment used for the specimen of Figures 27-29). After being quenched in water they were heated to 1500, 1600, and 1700 F for 100 hours. The specimen treated at 1500 F (Figure 31) shows blocky or globular carbides and a heavy precipitate throughout the grain, which might be γ', or, since it is overaged, Ni_3Cb. Treatment for 100 hours at 1600 F (Figure 32) produced an extraordinary Widmänstatten pattern of needles identified as Ni_3Cb. A background precipitate is also seen. In Figure 33, the structure obtained after 100 hours at 1700 F shows an agglomerated phase in addition to the Ni_3Cb needles. The agglomerated phase has been identified as Laves. As discussed earlier, in

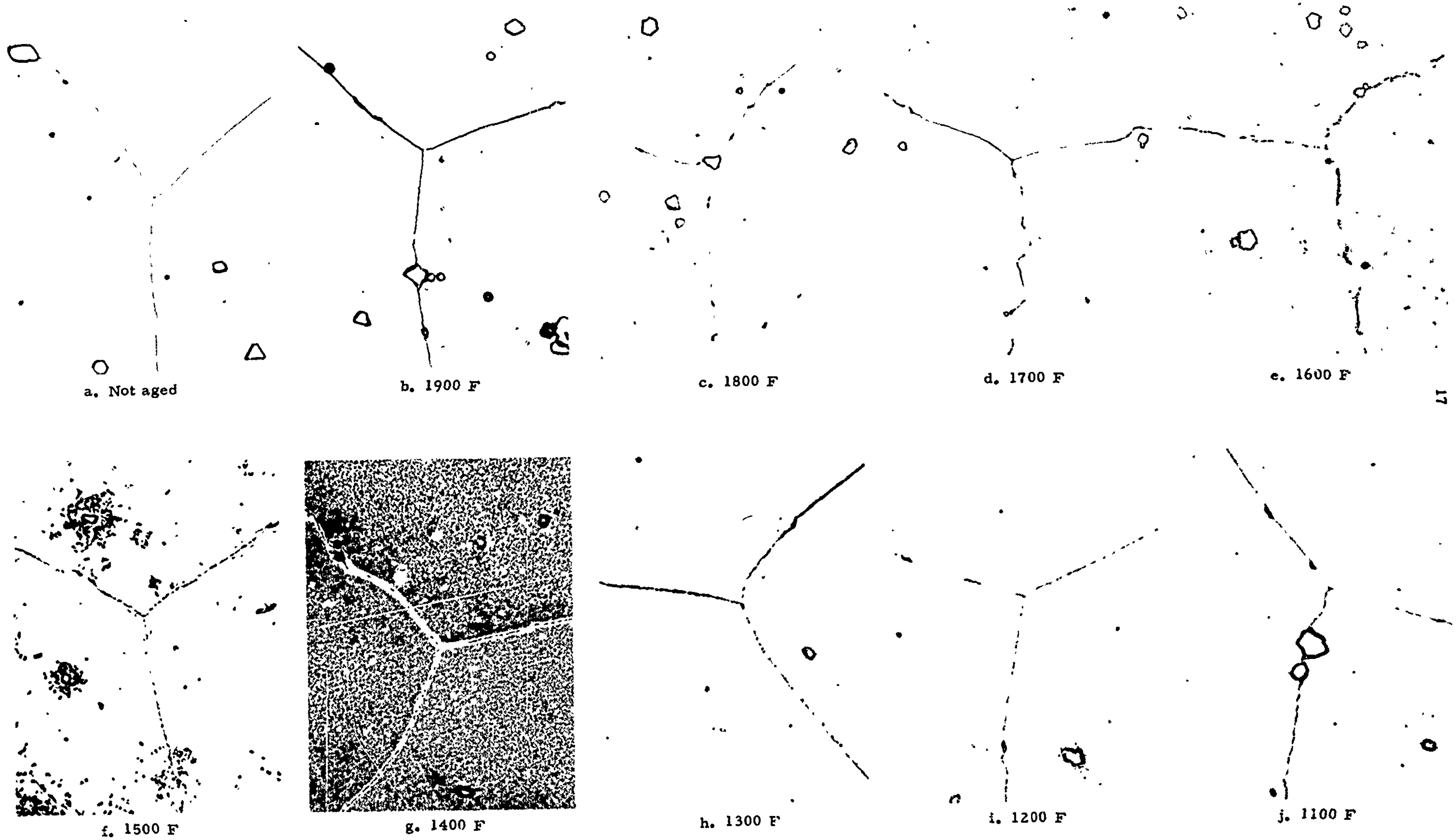
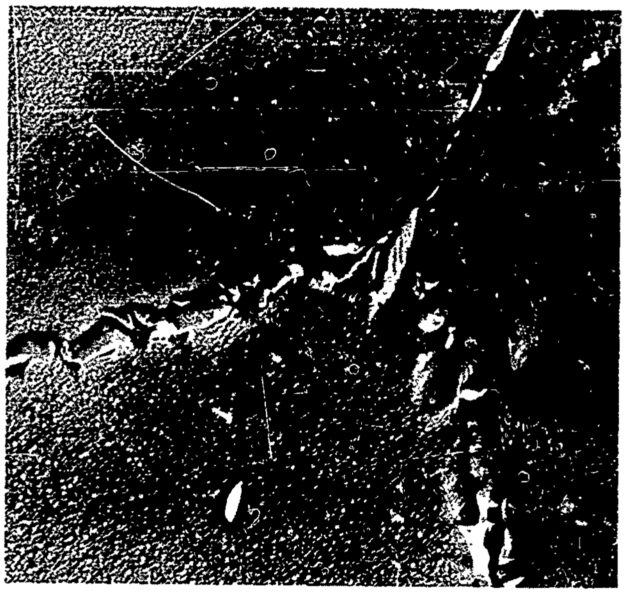


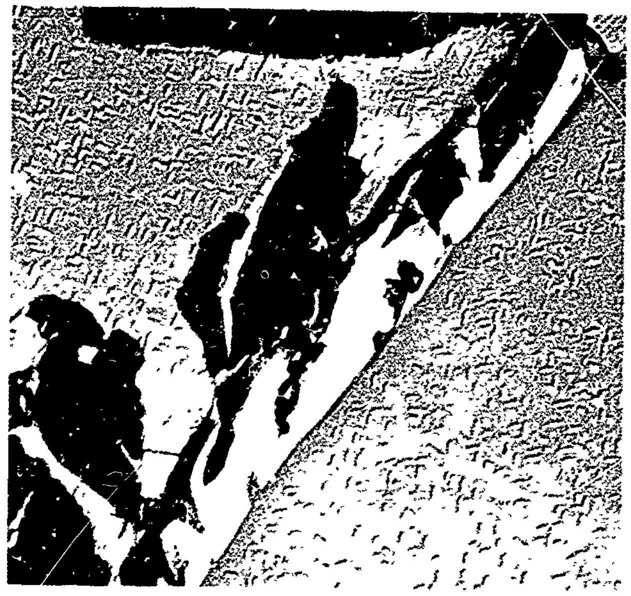
FIGURE 27. MICROSTRUCTURE OF HOT-ROLLED ALLOY 718 ANNEALED FOR 1 HR AT 2200 F AND HELD FOR 4 HR AT THE AGING TEMPERATURE INDICATED[22]
(Magnification 1500X. Etchant - 5 percent chromic acid. Reduced approximately 35 percent in printing.)



10,000 X

FIGURE 28. ALLOY 718 ANNEALED AT 2200 F FOR 1 HR, COOLED TO 1400 F, HELD 4 HR AND WATER QUENCHED(22)

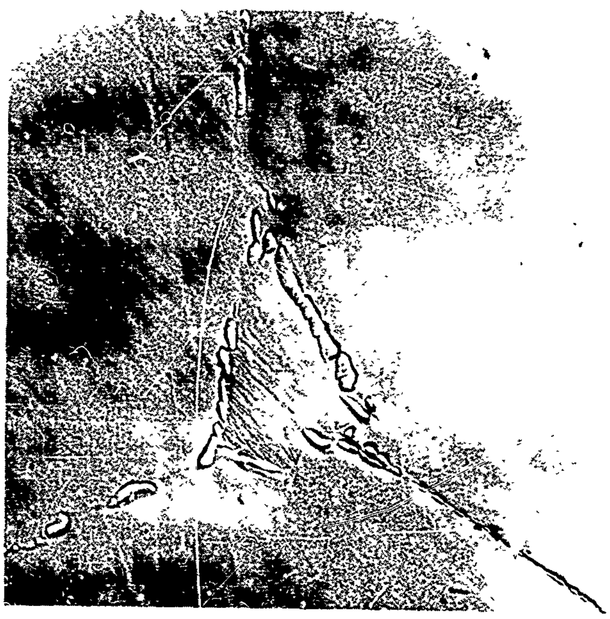
(Reduced approximately 20 percent in printing.)



17,500X

FIGURE 29. ALLOY 718 ANNEALED AT 2200 F FOR 1 HR, COOLED TO 1400 F, HELD 4 HR AND WATER QUENCHED(6)

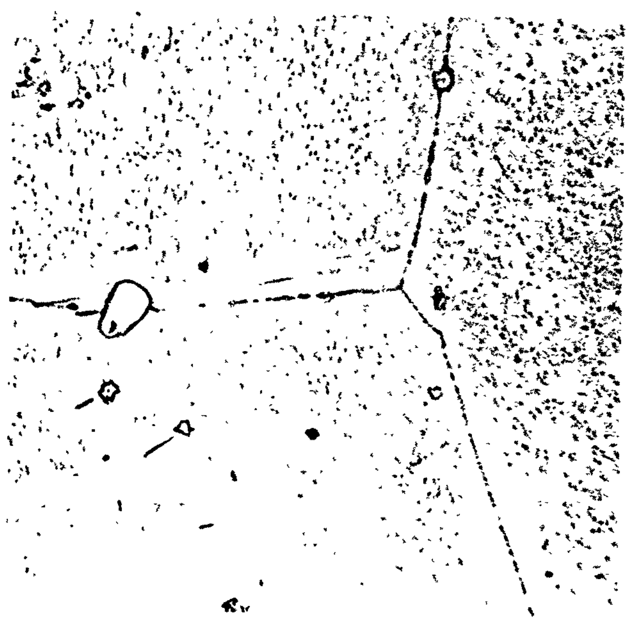
(Reduced approximately 20 percent in printing.)



5000 X

FIGURE 30. ALLOY 718 ANNEALED AT 2200 F FOR 1 HR, COOLED TO 1600 F, HELD 4 HR AND WATER QUENCHED(18)

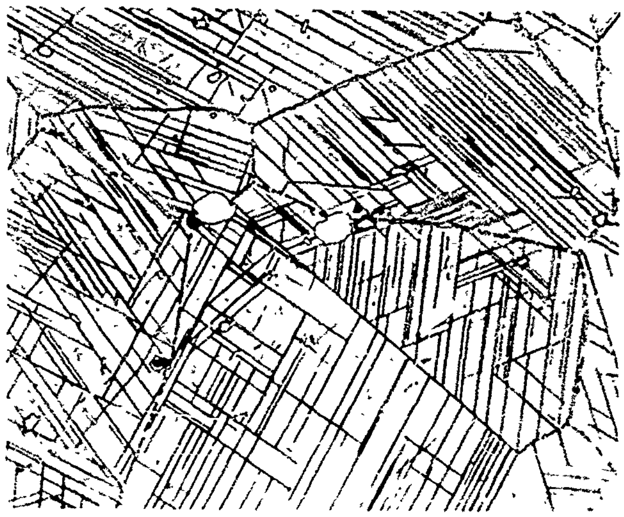
(Reduced approximately 20 percent in printing.)



1000 X 5% Chromic Acid Etch

FIGURE 31. ALLOY 718 ANNEALED AT 2100 F FOR 1 HR, WATER QUENCHED; AGED FOR 100 HR AT 1500 F, AIR COOLED(6)

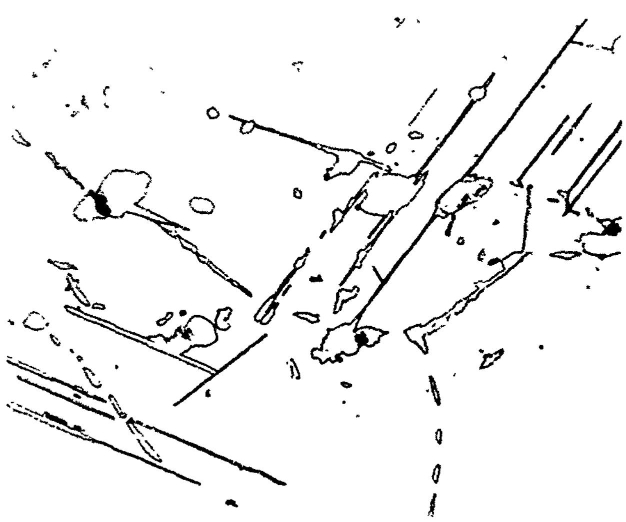
(Reduced approximately 20 percent in printing.)



1000X 5% Chromic Acid Etch

FIGURE 32. ALLOY 718 ANNEALED AT 2100 F FOR 1 HR, WATER QUENCHED; AGED FOR 100 HR AT 1600 F, AIR COOLED[6]

(Reduced approximately 20 percent in printing.)



1000X 5% Chromic Acid Etch

FIGURE 33. ALLOY 718 ANNEALED AT 2100 F FOR 1 HR, WATER QUENCHED; AGED FOR 100 HR AT 1700 F, AIR COOLED[6]

(Reduced approximately 20 percent in printing.)

connection with Laves phases of the A_2B type, it seems that long exposure to relatively high temperatures can promote the formation of this compound.

As Figure 34 shows, Kaufman and Palty[19] found the phase after 48 hours at 1700 F and after 24 hours at 1850 F.

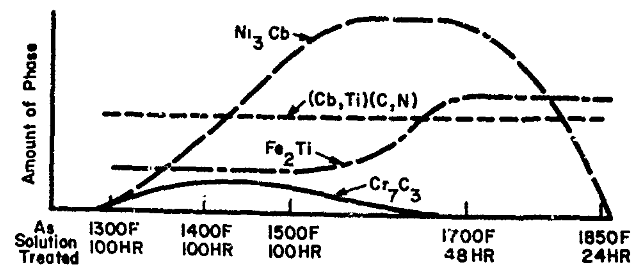


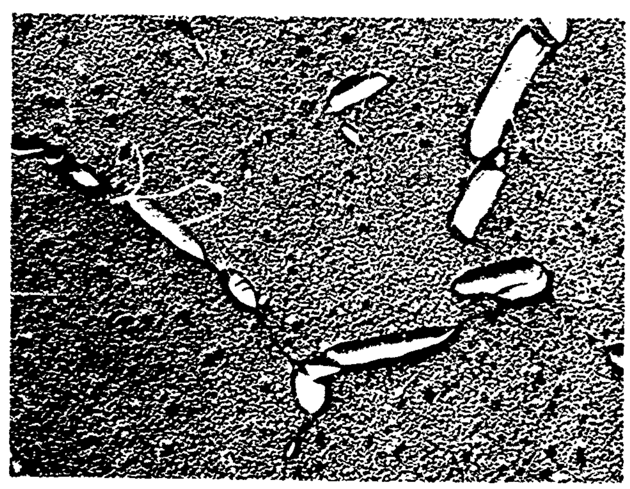
FIGURE 34. PHASES IN ALLOY 718 SOLUTION TREATED AT 2250 F AND AGED AS INDICATED[19]

Microstructures Obtained from Typical Heat Treatments and Service Conditions

Figures 35 and 36 are electron micrographs obtained from two different heat treatments.[18]

Figures 37-39 show the structures obtained after long-time relaxation tests at 1300 F and 50,000-psi initial stress. The specimens, annealed at different temperatures, all show over-aged gamma prime in the background, M_6C in the grain boundaries, and a few indications of orthorhombic Ni_3Cb needles. No Laves phase was detected.

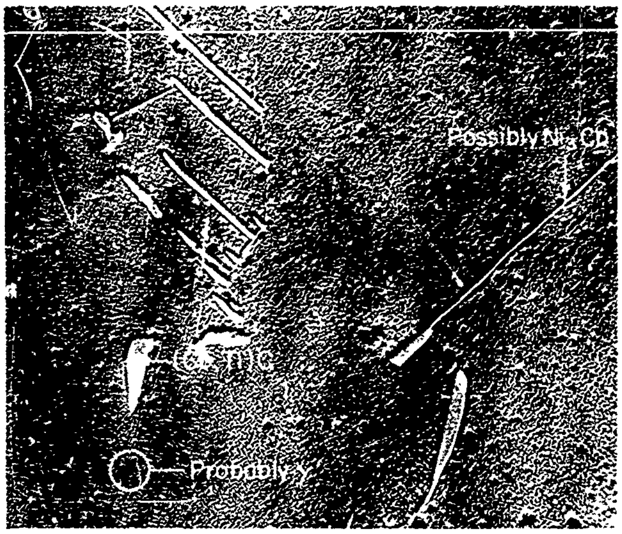
Finally, Figure 40 shows an electron micrograph of a stress-rupture specimen that had been tested for 5,470.7 hr at 1350 F and 10,000 psi.[6] Various phases are identified in the figure, which further illustrates the complexity of the alloy.



5000X

FIGURE 35. WROUGHT ALLOY 718 ANNEALED AT 1700 F, 1 HR, AIR COOLED; AGED AT 1325 F, 16 HR, AIR COOLED[18]

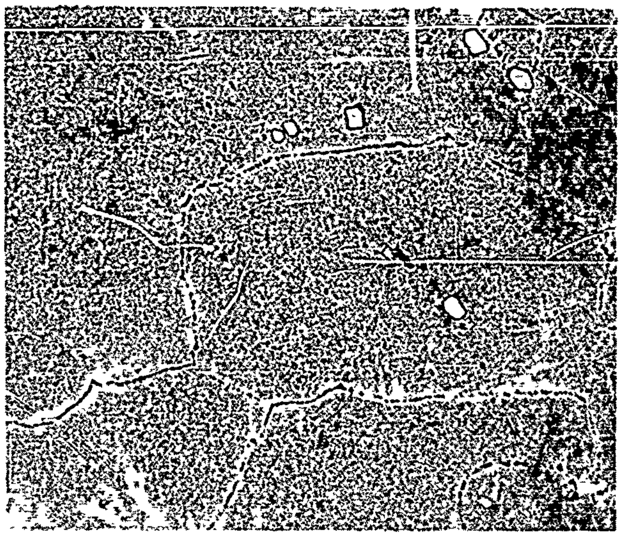
(Reduced approximately 20 percent in printing.)



5000X

FIGURE 36. WROUGHT ALLOY 718 ANNEALED AT 1700 F, 1 HR, AIR COOLED; AGED FOR 16 HR AT 1325 F, AIR COOLED; AGED AT 1200 F FOR 200 HR, AIR COOLED[18]

(Reduced approximately 20 percent in printing.)



1000X Chromic Acid Etch

FIGURE 37. ALLOY 718 AFTER RELAXATION TESTS AT 1300 F AND 50,000-PSI INITIAL STRESS FOR 1272 HR[6]

(Heat treatment: annealed for 1 hr at 1800 F, water quenched; aged at 1325 F for 16 hr, air cooled.)

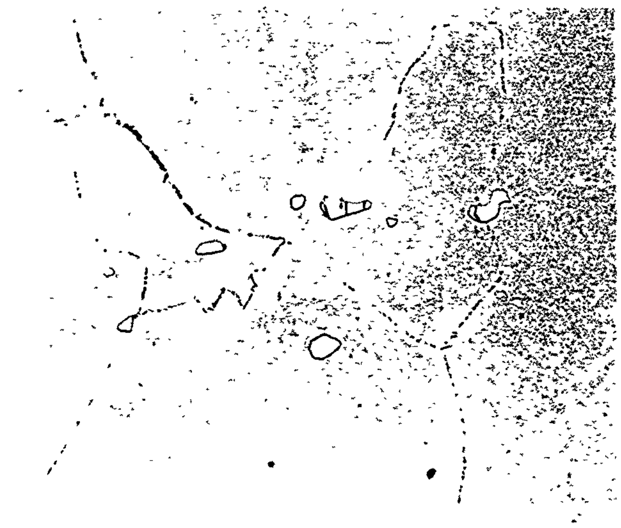
(Reduced approximately 20 percent in printing.)

1000X Chromic Acid Etch

FIGURE 38. ALLOY 718 AFTER RELAXATION TESTS AT 1325 F AND 50,000-PSI INITIAL STRESS FOR 1151 HR[6]

(Heat treatment: annealed for 1 hr at 1900 F, water quenched; aged at 1325 F for 16 hr, air cooled.)

(Reduced approximately 20 percent in printing.)

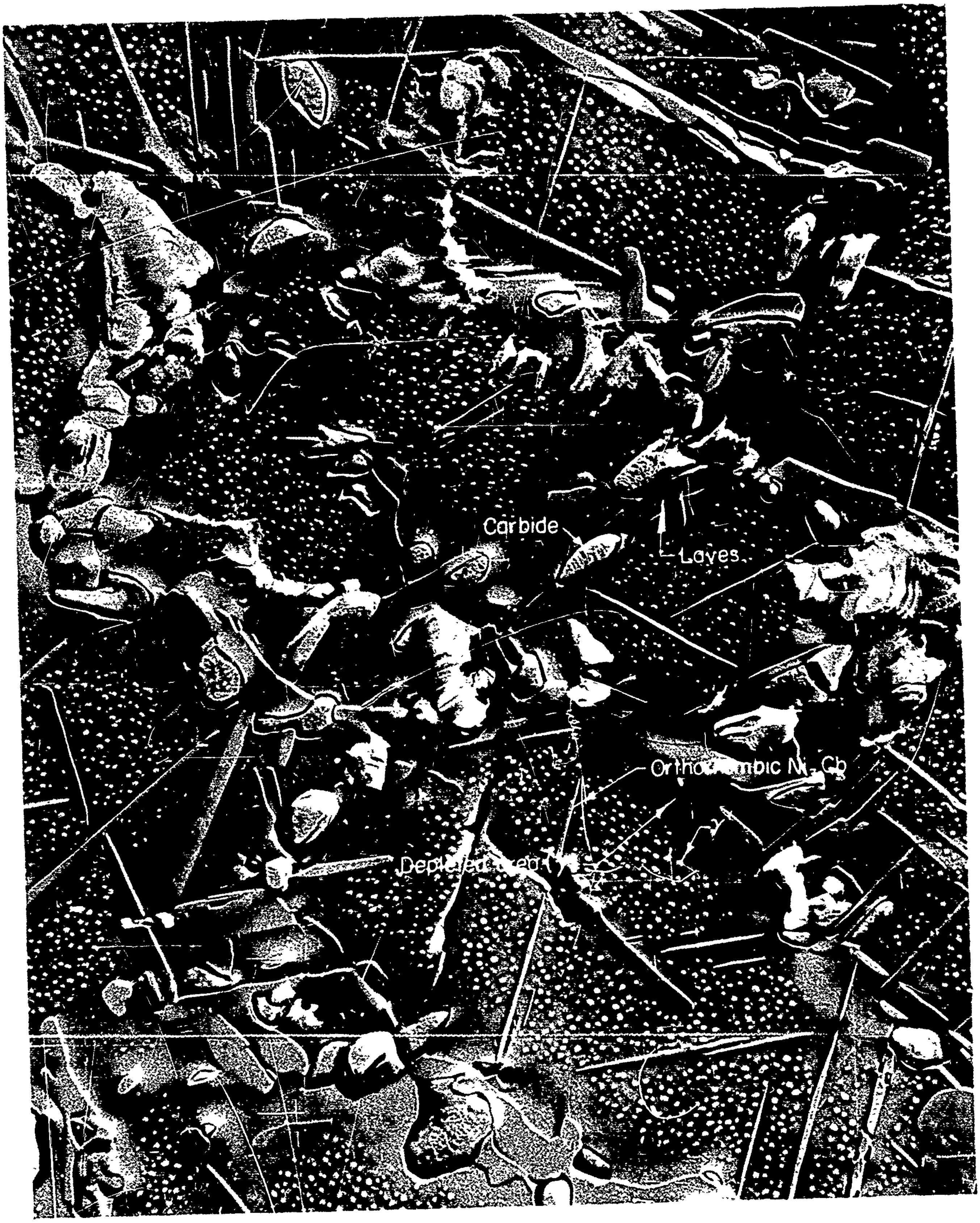


1000X Chromic Acid Etch

FIGURE 39. ALLOY 718 AFTER RELAXATION TESTS AT 1325 F AND 50,000-PSI INITIAL STRESS FOR 1583 HR[6]

(Heat treatment: annealed for 1 hr at 2000 F, water quenched; aged at 1325 F for 16 hr, air cooled.)

(Reduced approximately 20 percent in printing.)



10,000X

FIGURE 40. ALLOY 718 STRESS-RUPTURE SPECIMEN TESTED AT 1350 F, 10,000-PSI STRESS FOR 5470.7 HR(6)

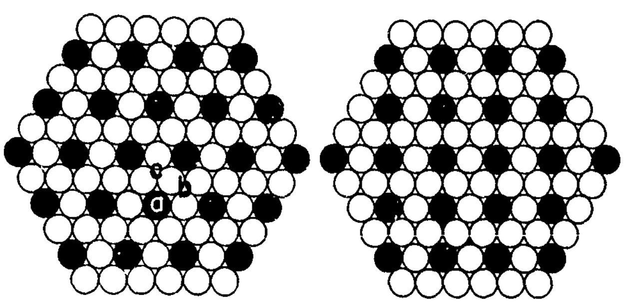
(Heat treatment: annealed at 1700 F, 1 hr, water quenched; aged at 1325 F, 16 hr, air cooled. Reduced approximately 20 percent in printing.)

STRENGTHENING MECHANISM

The crystallographic nature of the gamma prime (γ') constituent, and its role in strengthening Alloy 718, have been studied recently by Cometto.[23] The following summarizes his findings.

Gamma prime, as its name implies, is similar in many ways to the face-centered cubic matrix (γ) from which it forms. The only difference, in fact, is that γ' more nearly approaches the stoichiometric ratio A_3B, resulting in ordering of the atomic positions and a slight distortion of the lattice.

The A_3B-type intermetallic compounds can be classified according to the way the atoms are ordered. Figure 41 shows the two ways that close-packed planes can order so that each B-atom (black) has 6 A-atom (white) nearest neighbors.



Type A Layer　　Type B Layer

FIGURE 41. CLOSE-PACKED ORDERED LAYERS IN A_3B PHASES (After Nevitt)[24]

The Type A layers can occur in four different stacking sequences, and Type B layers in two different stacking sequences, giving six different crystal structures or families of compounds. Table 4 shows these compounds and the corresponding nickel intermetallic compounds.

The atoms of the Ni_3Al and Ni_3V compounds occupy essentially the same lattice sites as the atoms in the gamma solid solution. However, during the formation of γ', the atoms order on available sites in such a manner as to eliminate Al-Al or V-V nearest neighbors. Such a reaction (called an "exchange transformation") is characterized by a rapid and uniform nucleation. On the other hand, compounds such as Ni_3Ti (hexagonal structure) and Ni_3Cb (orthorhombic Cu_3Ti structure) require a complete rearrangement of atom sites as well as composition changes in order to precipitate from a face-centered cubic matrix. Accordingly, this transformation is much more difficult to nucleate than an exchange transformation.

TABLE 4. STACKING ARRANGEMENTS IN CLOSE-PACKED ORDERED A_3B STRUCTURES[23,24]

Structure Type	Nickel Compound	Layer Type	Stacking Sequence
Cu_3Au	Ni_3Al	A	abcabc...
Ni_3Ti	Ni_3Ti	A	abacabac...
Cd_3Mg	--	A	abab...
Al_3Pu	--	A	abcacbabcacb...
Cu_3Ti	Ni_3Cb	B	abab*
Al_3Ti	Ni_3V	B	abcdef*...

* Neglects slight distortion.

It was found that Alloy 718 precipitates a metastable γ' phase based on the Ni_3Cb composition, but with a body-centered tetragonal Ni_3V structure. Upon aging at 1400 F for 10 hours, furnace cooling at 100°/hr to 1200 F, holding 8 hours and air cooling, the lattice constants of the γ' were found to be

$$a_o = 3.624\text{Å}$$
$$c_o = 7.406\text{Å}$$
$$a_o/c_o = 2.044$$

Both the metastable Ni_3Cb gamma prime and the orthorhombic Ni_3Cb are made up of Type B layers, though apparently differing in stacking sequence. The transformation to γ' occurs by a simple rearrangement of atoms on existing lattice sites, and occurs rapidly and uniformly because it is not necessary to nucleate a new lattice. The lattice relationship between the γ' precipitate and the parent γ matrix was found to be:

$$[001]_{\gamma'} \| \langle 001 \rangle_{\gamma} \text{ and } \{100\}_{\gamma'} \| \{100\}_{\gamma}$$

The individual γ' particles are disc shaped (see Figure 32) and lie on the {100} matrix planes. The C_o axis of the γ' structure is perpendicular to the plane of the discs. This relationship results in three orientations (Figures 28 and 29) of γ' particles delineating three {100}-type gamma planes.

Cometto's analysis has shed considerable light on the γ' strengthening mechanism in Alloy 718. It can be used to explain why the double-aging treatment results in higher strength than the single aging. Apparently, to get maximum strengthening, it is necessary to precipitate as much γ' as possible, without overaging; that is, without transforming from the body-centered tetragonal γ' to the orthorhombic Ni_3Cb. High temperatures and long times favor the latter. Eiselstein[6] has plotted the conditions for forming γ' and Ni_3Cb as an isothermal transformation diagram. This is shown in Figure 42, with the double-aging treatment superimposed.

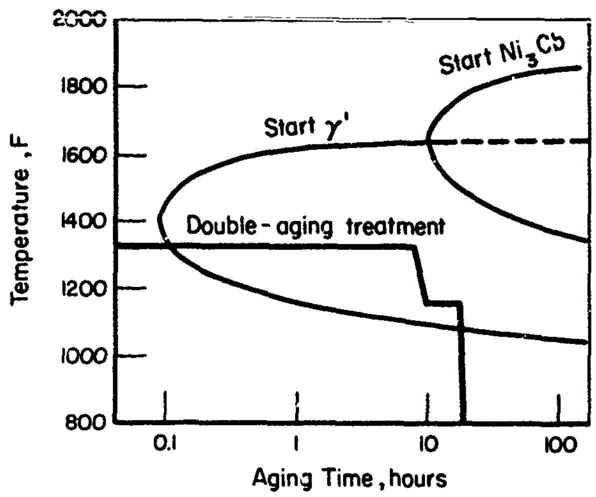


FIGURE 42. ISOTHERMAL-TRANSFORMATION DIAGRAM FOR GAMMA PRIME (γ') AND Ni_3Cb PHASES IN ALLOY 718; WITH DOUBLE-AGING TREATMENT SUPERIMPOSED (After Eiselstein)[6]

From Figure 42, it is seen that the treatment at 1325 F (or 1325-1425 F) would bring down maximum amounts of γ'. To be certain that no Ni_3Cb forms, the aging temperature is lowered to 1150-1200 F, which continues to precipitate γ' without transforming to the orthorhombic Ni_3Cb.

This explanation, though no doubt simplified, helps in understanding the relationships between heat treatment, properties, and microstructure. It does not take into account, however, the carbides, Laves phases, or variations in the diagram produced by chemical composition variations. Clearly, more investigation of these complex interactions needs to be carried out before a full understanding can be obtained.

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Unclassified
Security Classification

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13. ABSTRACT

This report deals with the physical metallurgy of Alloy 718. Since Alloy 718 was first introduced, it has broadened its areas of application. Primarily, it was intended for use at medium temperatures, 1300 F at most, its main applications being in lightweight welded frames and other assemblies in aircraft turbojet engines. Its good properties in the range of temperatures from -423 to 1300 F make it especially suitable for use in the LOX-LH_4 rocket engines. In these applications, Alloy 718 is used for the fuel/oxidizer injector plates, forged rings, thrust chamber jackets, turbine wheels, bellows, and tubing. This report presents information on the effect of composition and heat treatment on mechanical properties; trends in specifications for Alloy 718; microstructure and microconstituents of the alloy and the strengthening mechanism. No attempt has been made to present a complete compilation of design properties in this report.

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14. KEY WORDS	LINK A		LINK B		LINK C	
	ROLE	WT	ROLE	WT	ROLE	WT
Physical Metallurgy	8	3				
Microstructure	8	3				
Strengthening Mechanism	8	3				
Alloy 718	8,9	3	9	3		
Frames	4	3				
Engines	4	3				
Aircraft	0					
Turbojet	0					
Rocket	0					
LOX-LH_2	0					
Injector plates	4					
Forged rings	4	3				
Thrust chamber jackets	4	3				
Turbine wheels	4	3				
Bellows	4	3				
Tubing	4	3				
Composition			6	3		
Heat treatment			6	3		
Mechanical properties			7	3		
Specification	8					

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