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A REVIEW OF THE PHYSICAL METALLURGY
OF ALLOY 718

MASTER

PREPARED BY AEROJET NUCLEAR COMPANY FOR
ENERGY RESEARCH AND DEVELOPMENT ADMINISTRATION
IDAHO OPERATIONS OFFICE UNDER CONTRACT E(10-1) -1375

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**A REVIEW OF THE PHYSICAL METALLURGY
OF ALLOY 718**

by

D. D. Keiser
H. L. Brown

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ABSTRACT

The physical metallurgy of Alloy 718 is updated to 1976 on the basis of a survey of post-1967 literature and current experimental data. Composition, microstructures, and mechanical properties are correlated with heat treatment parameters. The current state of understanding of phase stability, strengthening mechanisms, deformation modes, and recovery and recrystallization in this material is described.

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A REVIEW OF THE PHYSICAL METALLURGY OF ALLOY 718

I. INTRODUCTION AND SUMMARY

Alloy 718 is a precipitation-hardenable material, used primarily in the wrought product form. It has good tensile properties at cryogenic, ambient, and intermediate temperatures and exhibits excellent weldability for a high strength material. Its major uses have been for aircraft turbine engines, rocket motors, and, lately, in nuclear reactors. In 1966, it was one of the largest tonnage superalloys; around one million pounds were produced. It was estimated^[1] that about seven million pounds of this material would be produced in 1972. The International Nickel Co. developed this material in the late 1950's. H.L. Eiselstein was primarily responsible for its development for International Nickel Co. The primary aim^[2] of the International Nickel Co. in the initial development of this material was to achieve high tensile properties and stability of the age-hardening constituents.

The work reported here represents an update of the work of R.C. Hall^[3], whose survey article, "The Metallurgy of Alloy 718", compiled the available information on this material up to 1967. This work was reviewed, and more recent efforts were surveyed^[4].

Additional information from the literature and new experimental data^[4] were compiled to update the state of the art for this material to 1975. The overall picture has been put into perspective by examining and correlating composition, heat treatment variables, mechanical properties, and microstructures. The factors which affect phase stability have been studied, strengthening mechanisms and deformation modes have been reevaluated, and recovery and recrystallization have been characterized. The results of these investigations and evaluations are presented and discussed under the various subject headings.

This report describes and documents, in some detail, the current understanding of the physical metallurgy of Alloy 718. Much of the basic information and interpretation for this material is equally applicable to other high-nickel, precipitation-hardening alloys.

II. RESULTS AND DISCUSSION

1. COMPOSITION

Presented in Table I is the nominal composition of Alloy 718 along with the nominal compositions of several other superalloys; the other superalloys are presented for comparison purposes.

The effects of the individual elements^[3] on properties will now be presented, where possible.

1.1 Nickel

The effect of nickel content on yield strength and stress rupture strength is shown in Figure 1. Basically, it shows that at approximately 53% nickel, maximum yield strength is obtained. An increase in nickel content generally improves stress rupture properties but only at the expense of strength above 53% nickel.

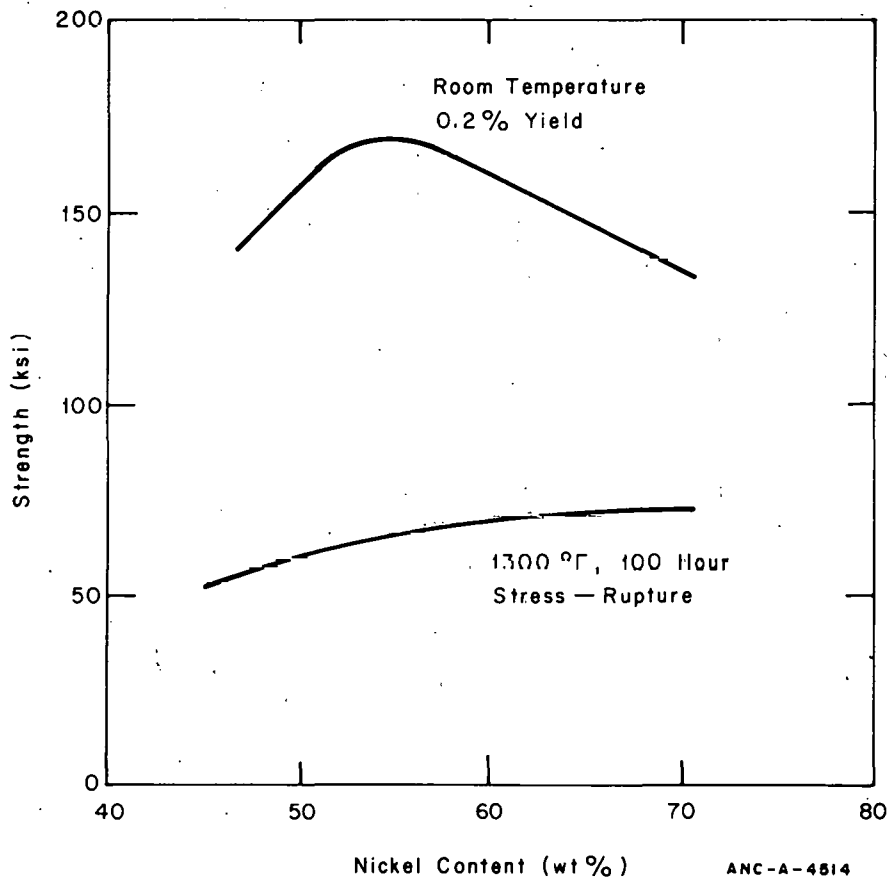


Fig. 1 The effect of nickel content on yield strength and stress rupture strength of Alloy 718^[2].

TABLE I
COMPOSITION OF ALLOY 718 AND SEVERAL OTHER SUPERALLOYS^[3]

<u>Alloy</u>	<u>Base</u>	<u>Composition (w/o) For Element</u>										
		<u>C</u>	<u>Cr</u>	<u>Ni</u>	<u>Cb</u>	<u>Mo</u>	<u>Ti</u>	<u>Al</u>	<u>B</u>	<u>Fe</u>	<u>Co</u>	<u>Others</u>
718	Ni	0.05	19	53	5	3	1	0.5	0.005	18	-	-
Rene' 41	Ni	0.09	19	53	-	10	3	1.5	0.005	-	11	-
Waspaloy	Ni	0.08	19	56	-	4.3	3	1.3	0.006	-	13.5	-
A-286	Fe	0.05	15	26	-	1.3	2	0.2	0.015	54	-	-
S816	Co	0.38	20	20	4	4	-	-	-	4	43	4

1.2 Chromium

Chromium is generally added to provide oxidation resistance and to provide solid solution hardening of the matrix (nickel and iron). In order not to reduce hot workability, chromium is seldom employed in an amount greater than 19 wt%.

1.3 Molybdenum

Molybdenum is added for high temperature strength through solid solution hardening. Apparently, it does not precipitate from solution at moderate temperatures but remains in solid solution. Like chromium, it too can impair hot workability if added in large amounts; it is kept at 3 wt% or less to assure that this does not happen.

1.4 Columbium

Strength through precipitation hardening can be increased significantly by the addition of columbium. This is illustrated in Figure 2. However, concurrent with the large increase in strength is a significant drop in ductility (see Figure 2), especially with amounts greater than 5 wt%. Therefore, the optimum columbium content is established at 5 wt% so that good ductility with average strength can be obtained.

1.5 Titanium

Titanium, like columbium, is one of the constituents involved with the age-hardened precipitate in the gamma matrix. It can significantly raise strength but at the same time reduce ductility. The reaction of titanium is not simply described^[3] since it appears to react with both aluminum and carbon to give variable effects. Heat treatment variables also greatly affect the impact of titanium on properties. The established 1% level is an optimum composition for obtaining good tensile strength and stress rupture properties. Eiselstein^[2] shows that stress rupture strength is a maximum with 1% titanium.

1.6 Aluminum

The titanium and carbon content, as well as the heat treatment variables, all affect the response of mechanical properties to the aluminum content. Aluminum, like titanium, is a constituent of the precipitation-hardening phase. It is reported^[3] that above the nominal 0.5% Al level, the strength is decreased. Aluminum does not seem to improve ductility and stress rupture life. In general, the precise effect that aluminum has on properties has not been fully explained^[3].

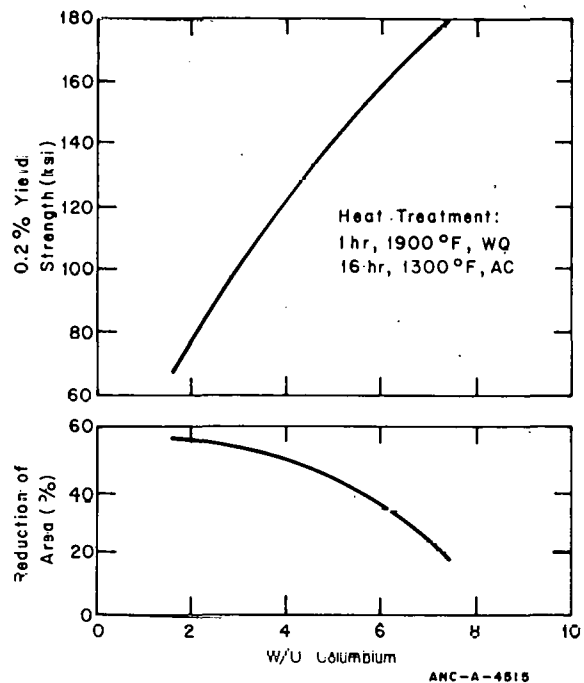


Fig. 2 The effect of columbium content on room temperature yield strength and ductility^[5].

1.7 Carbon

Carbon shows a great tendency to react with columbium to form CbC. This reaction robs columbium from the precipitation-hardening phase and in effect reduces strength. Therefore, carbon is kept low, approximately 0.05%. Some people feel that carbon present in the form of CbC may be helpful in maintaining a fine grain size during high temperature heating operations^[3,5].

1.8 Other Elements

Small additions of boron are helpful in improving stress rupture properties; therefore, 0.005% has been added, an optimum based on regression analysis by Eiselstein^[2].

Silicon must be kept low to avoid reducing hot workability and impairing stress rupture properties^[2]. It is thought to embrittle the solid solution matrix^[3].

Nitrogen, which occurs as TiN, should be minimized because as carbon combines with columbium and causes strength reduction, nitrogen combines with titanium and effectively removes it as a potential hardening agent.

2. TYPICAL MECHANICAL PROPERTIES

Figure 3 gives the comparative mechanical properties of Alloy 718 and other superalloys at elevated temperatures. It shows Alloy 718 to have good high temperature strength with moderate ductility and comparatively high stress rupture properties. Strength drops off fast, however, beyond 1,400°F (760°C).

3. MICROSTRUCTURES

Over the years, a large number of different phases have been thought to exist in Alloy 718. This number has been reduced in the last few years with the advent of better qualitative test methods with which to identify these phases. Ideally, in the fully age-hardened conditions, Alloy 718 should contain gamma (the matrix material), gamma prime (γ'), gamma double prime (γ''), and a small amount of CbC and TiN. However, because of segregations from casting, improper heat treatments, or high temperature exposure in service, other phases appear. These are Laves, Ni_3Cb , and high temperature carbides. Each of these phases will be discussed with regard to such factors as approximate chemistry, morphology, crystal structure, formation characteristics, and effect on mechanical properties.

3.1 Gamma Phase

Gamma is the matrix material and consists of the solid solution elements: nickel, chromium, iron, columbium, molybdenum, titanium, and aluminum. It has a face-centered cubic structure and is comparatively weak in its solution-annealed condition (40 to 60 ksi

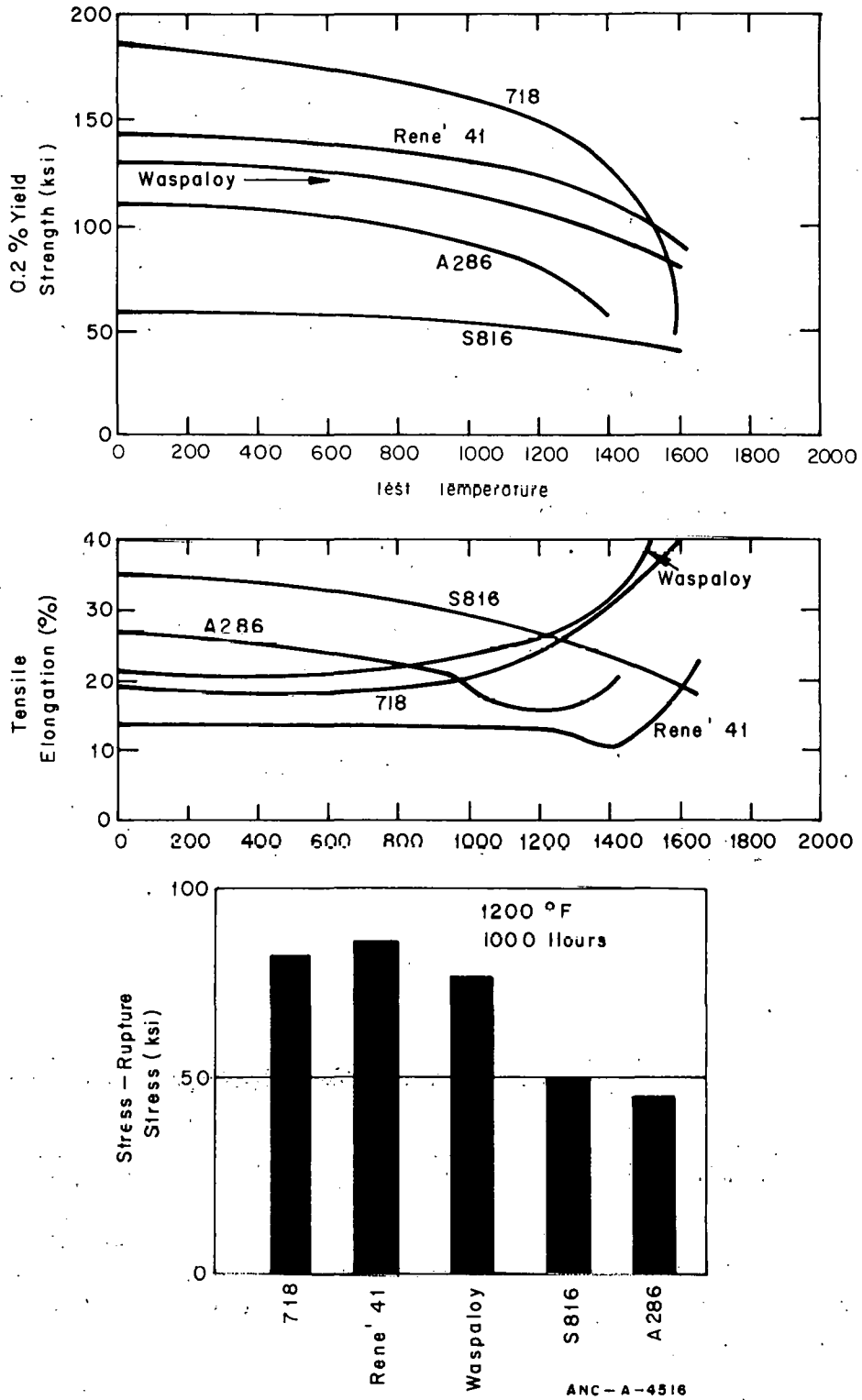


Fig. 3 Elevated temperature mechanical properties of Alloy 718 and other superalloys [6, 7].

yield strength). Transmission electron microscopy (TEM) studies^[8] have shown that a supersaturated solid solution of gamma can be maintained on air cooling from 1,750°F (954°C).

3.2 Gamma Prime (γ') Phase

Gamma prime is an A_3B compound where the A atoms are primarily nickel, and the B atoms are largely columbium with some Ti and Al^[2]. Eiselstein^[2] reports the approximate chemical composition of gamma prime to be, in atomic percent: 75% nickel, 15% columbium, 5% titanium, and 5% aluminum.

Gamma prime is an ordered fcc ($L1_2$) structure. The unit cell structure, shown in Figure 4, consists of Cb, Al, and Ti atoms at the corners and Ni atoms at the center of the faces.

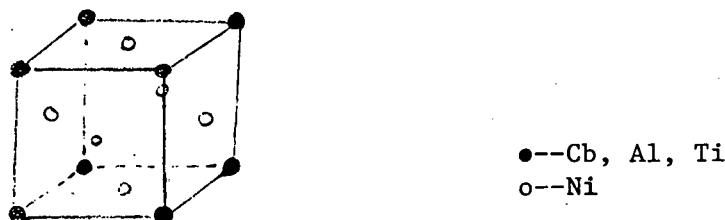


Fig. 4 Unit cell structure of gamma prime.

Gamma prime has a spherical shape and appears as small dots within the gamma matrix. It appears to form uniformly through the matrix during aging. It is coherent with the matrix and begins to form at 1,200°F (649°C) within reasonable periods of time (50 minutes) and has a solvus between 1,550 and 1,600°F (843 and 871°C)^[8]. The most rapid phase formation is at 1,350 to 1,400°F (732 to 760°C)^[2,5].

It was assumed for several years that gamma prime was the primary hardening constituent in Alloy 718^[2]. More recent studies show that gamma double prime may be the primary hardening constituent (more will be said on this later). However, gamma prime does contribute to the hardening of Alloy 718, but only 10 to 20%^[8]. It also plays a major role in the strengthening mechanism by inhibiting deformation by a stacking fault mode of shear^[9]. Many people feel that γ' transforms to γ'' upon exposure at temperatures around 1,450 to 1,500°F (788 to 816°C) for one hour^[2,10,11].

3.3 Gamma Double Prime (γ'') Phase

Gamma double prime (γ''), like gamma prime, is an A_3B compound; however, it has an ordered body-centered-tetragonal structure, which is illustrated in Figure 5. In γ'' , as shown in Figure 5, columbium atoms are located at the corners and in the center of a cell formed by placing two fcc cells one upon the other^[8].

●-- Cb, Al, Ti
 ○-- Ni

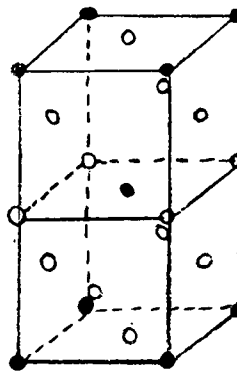


Fig. 5 Unit cell structure of gamma double prime.

The cell structure is actually a bct (DO_{22}) structure which is metastable and exhibits only a slight distortion ($c/a = 2.04$); therefore, it represents an ordered variation of the fcc lattice. It has been shown that it transforms into Ni_3Cb (orthorhombic) after exposure at temperatures above $1,500^{\circ}F$ ($816^{\circ}C$)^[9,10].

The γ'' precipitate forms as discs on the (100) planes with an average diameter of 600 \AA (Angstrom units) and a thickness of 50 to 90 \AA with normal exposures at temperatures between $1,200$ to $1,400^{\circ}F$ (649 to $760^{\circ}C$)^[8]. Above $1,400^{\circ}F$ ($760^{\circ}C$), it exhibits accelerated coarsening.

At present, there is little doubt that the strengthening phase in Alloy 718 is the metastable body-centered-tetragonal (DO_{22}) Ni_3Cb precipitate rather than the ordered fcc ($L1_2$) γ' . Therefore, anything that influences the amount, morphology, transformation, etc., of the phase will have an effect on mechanical properties.

3.4 Ni_3Cb Phase

This phase is an A_3B compound that has an ordered orthorhombic structure^[5]. The A atoms of the compound are primarily nickel atoms, and the B atoms are primarily columbium atoms with some aluminum and titanium. Boesch et al^[10] give the composition as $Ni_3(Cb_{0.8}Ti_{0.2})$.

Ni_3Cb forms as large needles in the grains and at the grain boundaries. Although Ni_3Cb is considered to be quite stable, it is not coherent with the matrix. Therefore, it contributes little or nothing to the strength and may harm ductility^[2,3]. It should be avoided where possible.

At temperatures above $1,800$ or $1,900^{\circ}F$ (982 or $1,038^{\circ}C$)^[3], Ni_3Cb generally dissolves. There are differences in opinions as to how long it takes Ni_3Cb to form in the $1,400$ to $1,800^{\circ}F$ (760 to $982^{\circ}C$) temperature range. Eiselstein^[2] reports that at $1,700^{\circ}F$ ($927^{\circ}C$), it begins to form after five hours. However, Boesch and Canada^[10] report that it begins to form after only 15 to 20 minutes at $1,700^{\circ}F$ ($927^{\circ}C$). Eiselstein^[2] shows that the start of formation of Ni_3Cb needles is affected by the prior solution-anneal temperature; the effect of prior solution-anneal temperature is illustrated in Figure 6.

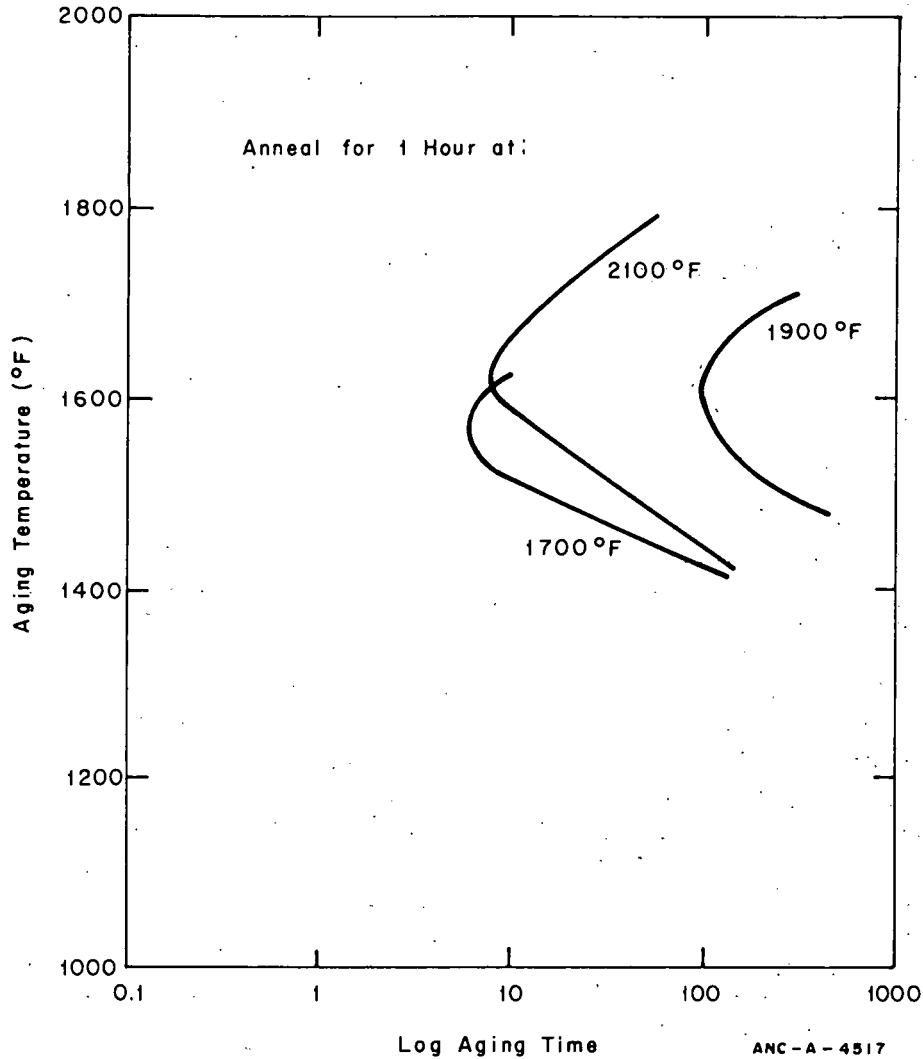


Fig. 6 Isothermal transformation diagram for start of formation of Ni_3Cb needles in Alloy 718^[2].

3.5 Laves Phase

Laves phase is an A_2B compound which is identical in structure to Fe_2Ti . The Laves phase consists of nickel (40 to 45%), columbium (25%), iron (10%), chromium (10%), molybdenum (10%), and possibly titanium^[3]. The A atoms are probably Ni, Fe, and Cr while the B atoms are probably Cb, Mo, and Ti.

This phase appears as round massive particles and is found in heavily corrod or segregated areas^[2]. Under the microscope, the phase appears as bright white particles.

Laves phase can form during solidification and during heat treatment. During solidification, it forms interdendritically in segregated areas at relatively low temperatures of 2,100 to 2,200°F (1,149 to 1,204°C)^[2]. During heat treatment, at times to ten hours, it

can form at low temperatures of 1,450 to 1,550°F (788 to 843°C). The phase is definitely harmful to mechanical properties in that it impairs ductility. In addition, it removes alloying elements required for precipitation hardening and therefore tends to reduce strength. It is removed by hot working at temperatures above 2,000°F (1,093°C)^[3].

Laves phase can form at grain boundaries, especially in weldments. For this reason, weldments must be postweld annealed at 2,000°F (1,093°C) to remove it if good weldment ductility is needed.

3.6 CbC and TiN Phases

TiN is apparently of little concern^[3] in Alloy 718 since it forms in such small amounts during solidification. It appears under the microscope as a pink phase, massive and angular in shape. Barker^[12] reports that it is stable up to the melting point.

CbC, however, forms in larger quantities and is apparently not as stable as TiN^[3]. It forms on solidification as massive angular particles. These carbides, which appear gray under the microscope, have a tendency to remove columbium and thus reduce strength. It can form in large amounts at grain boundaries^[2].

CbC dissolves on annealing above 1,900°F (1,038°C) and will precipitate on aging from 1,300 to 1,800°F (704 to 982°C) as a grain boundary film^[12]. Eiselstein^[2] did not find that its presence affected mechanical properties; however, by the very fact that it can be a grain boundary film, it should be avoided. Eiselstein does show that formation of CbC film can be pronounced upon aging after annealing at 2,100°F (1,149°C) where CbC is dissolved. On annealing at 1,700°F (927°C), the film should not be precipitated on aging because none is dissolved^[3]. The effect of annealing at temperatures above 1,700°F (927°C) on CbC film formation is illustrated in Figure 7 where it is shown that with only a prior anneal of 1,900°F (1,038°C), an aging time of about ten hours at 1,400°F (760°C) is required for a CbC film to form; and with an anneal of 2,100°F (1,149°C), only a few minutes are required for the film to form. No CbC film will form upon aging with a 1,750°F (954°C) solution anneal.

3.7 M₆C Phase

This form of carbide phase is considered to be harmless because of the small amounts present and the form in which it precipitates. It forms sluggishly compared to other phases discussed and results from long time aging at temperatures from 1,300 to 1,800°F (704 to 982°C). It can form in grain boundaries^[2].

3.8 Banding

Banding, which is segregation into alternate layers of secondary phases, is observed in Alloy 718 quite frequently, especially in plate material. Wagner et al^[5] feel that it results from interdendritic solidification of secondary phases in the form of rods which elongate during hot working. Bands can be dissolved above 1,800°F (982°C), which suggests that at least one constituent of these bands might have a chemistry similar to stable Ni₃Cb. Indeed, chemical analysis of the bands seems to indicate that they are a combination of stable Ni₃Cb and a high temperature carbide. An example of banding is shown in Figure 8. There are no documented accounts where banding has been harmful to mechanical properties. However,

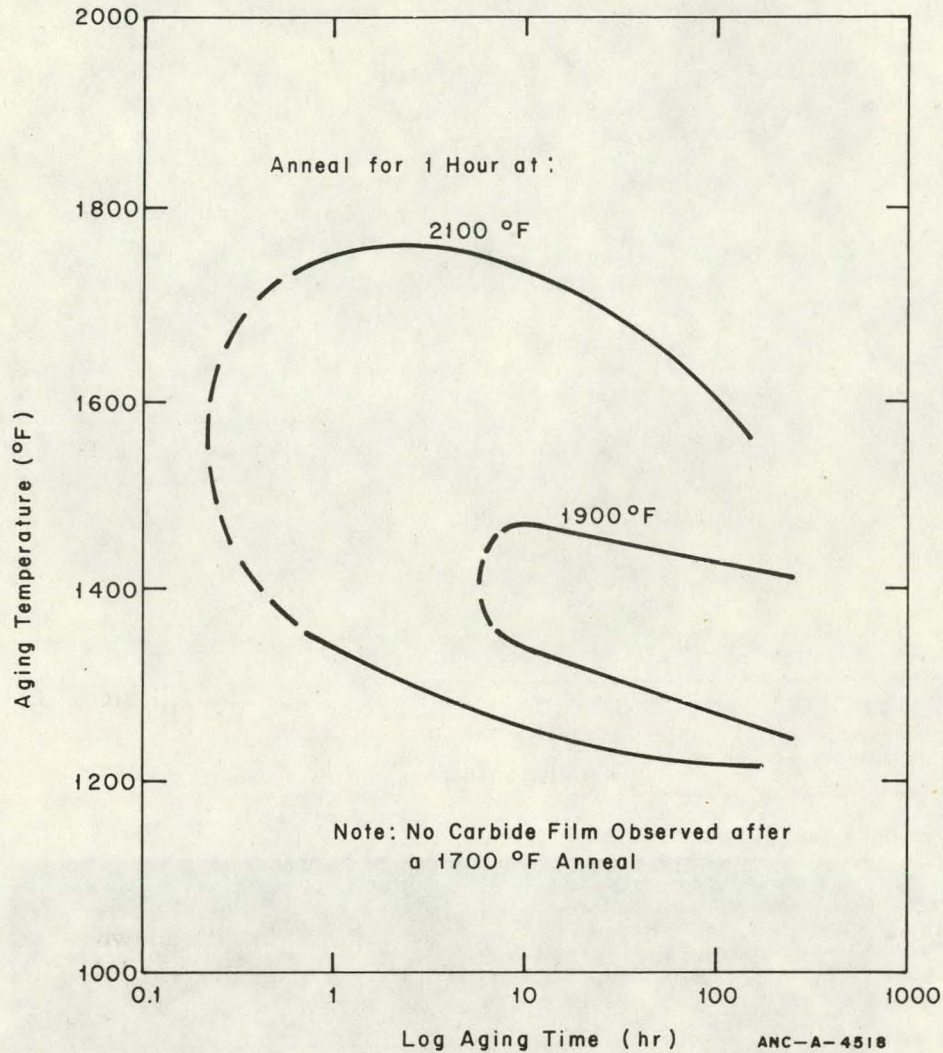


Fig. 7 Isothermal transformation diagram for start of formation of grain boundary CbC film in Alloy 718[2].

as shown in Figure 9, at least along its boundary, it can be a preferred path for crack propagation; and conceivably, under the right stress conditions (such as normal to its lamellar orientation in plate) and given amounts, it could be harmful to mechanical properties. It should be avoided where possible. Hot working above 1,800°F (982°C) should prevent its occurrence in wrought products.

3.9 Examples of Microstructures

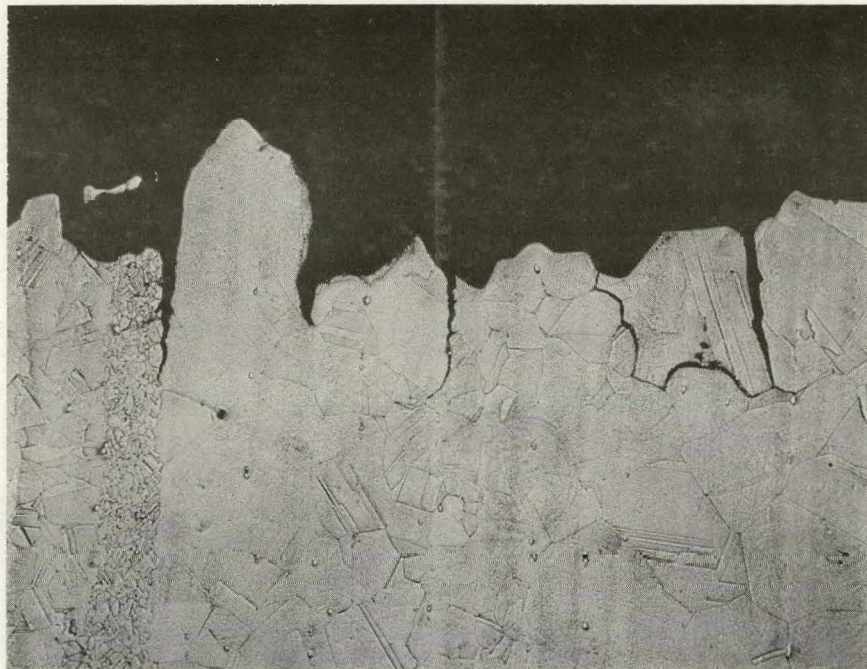
Presented in Figures 9 through 16 are examples of the phases described above. The examples are from carbon replication and transmission electron microscopy studies on material tested in various heat treat conditions.

Figures 10 and 11 show two conditions of gamma prime: (a) a hyperfine condition (Figure 10), and (b) a coarsened or agglomerated condition (Figure 11). In both situations, it shows the spherical particle shape typical of gamma prime. The small dark spherical particles in Figure 11 might be the product of nucleation of gamma double prime. Figure 12



(100X)

Fig. 8 Banding in Alloy 718.



(100X)

Fig. 9 Cracking along banded structure.

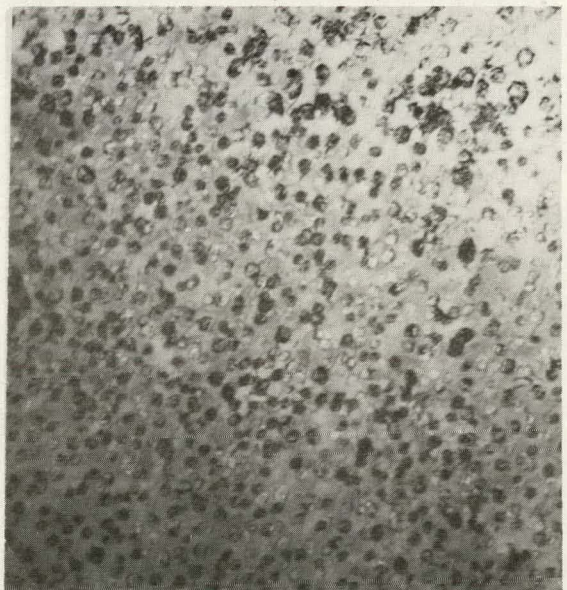
is a fine example of the disc-shaped gamma double prime. Both gamma prime and gamma double prime can be observed in Figure 13. It is interesting to note the frequent contiguous arrangement of gamma double prime to the gamma prime particles. This observation is in agreement with observations made by Cozar and Pineau^[13]. They use this observation for their argument that gamma double prime precipitates on gamma prime particles and that, at low aging temperatures, gamma double prime is always associated with gamma prime.

Ni_3Cb needles at grain boundaries can be observed in Figure 14. Also, the possible existence of Laves phase (massive round particles) can be seen in the lower right portion of the micrograph in Figure 14. It is highly possible that these particles are Laves phase because the material was heated only to $1,750^\circ\text{F}$ (954°C), a temperature at which Laves phase will not dissolve. CbC particles and possibly M_6C particles at a grain boundary can be seen in Figure 15. Information reported earlier would seem to indicate that they are predominately CbC particles. An extremely interesting phenomenon can be seen in Figure 16 where a classical case of grain boundary denudation has been detected in material heated to $2,000^\circ\text{F}$ ($1,093^\circ\text{C}$) and air cooled to room temperature. Associated with this is an apparent grain boundary film which is probably CbC . The phenomenon of grain boundary denudation has been observed by others^[14,15,16] and is apparently caused by CbC precipitation in the grain boundaries. When this happens, the carbon in the reaction robs columbium from the matrix material adjacent to the grain boundaries and prevents the precipitation-hardening constituents, consisting mainly of Ni_3Cb , from forming near the grain boundaries.



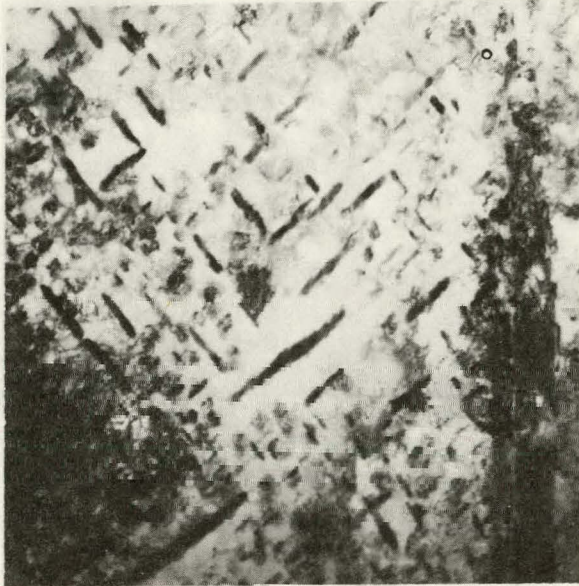
(30,000X)

Fig. 10 TEM of material heated to $2,000^\circ\text{F}$ ($1,903^\circ\text{C}$) and air cooled, showing hyperfine gamma prime.



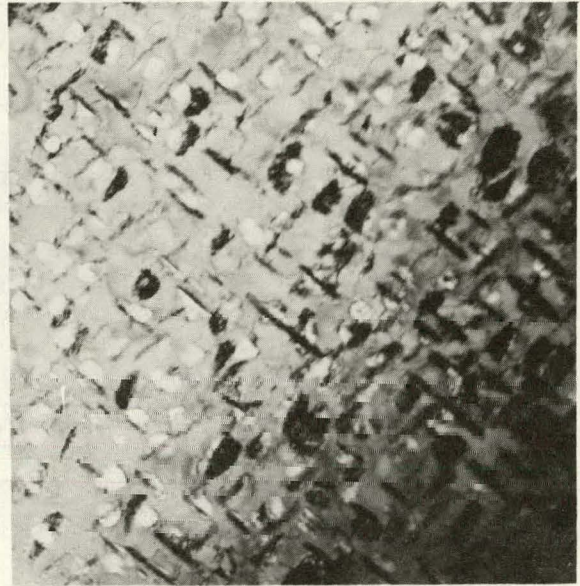
(30,000X)

Fig. 11 TEM of material heated to $2,000^\circ\text{F}$ ($1,093^\circ\text{C}$) and cooled 100°F (55°C)/hr, showing agglomerated gamma prime.



(30,000X)

Fig. 12 TEM of material heated to 2,000°F (1,093°C) and cooled 100°F (55°C)/hr, showing gamma double prime.



(30,000X)

Fig. 13 TEM of material heated to 2,000°F (1,093°C) and cooled 100°F (55°C)/hr, showing both gamma prime and gamma double prime.



(10,000X)

Fig. 14 Carbon replication showing Ni₃Cb needles and possible Laves phase in material heated to 1,750°F (954°C) and cooled at 100°F (55°C)/hr.



(10,000X)

Fig. 15 Carbon replication showing possible CbC and M_6C at grain boundary in material heated to 2,000°F (1,093°C) and air cooled.



(30,000X)

Fig. 16 TEM of material heated to 2,000°F (1,093°C) and air cooled, showing grain boundary denudation and possible CbC film in grain boundary.

4. PHASE STABILITY

In this section, discussions will be confined to the behavior of phases in Alloy 718 under conditions of time and temperature, chemistry, and thermomechanical effects.

4.1 Time-Temperature Effects

The most current^[17] time-temperature-transformation (TTT) diagram on Alloy 718 is shown in Figure 17. Although there is considerable disagreement^[2,9,10], it now appears that the precipitating phases follow the sequence of $\gamma' \rightarrow \gamma'' \rightarrow \text{Ni}_3\text{Cb}$. The region of $\gamma' \rightarrow \gamma''$ is under debate^[17] and is not well defined. There is some disagreement whether γ' must exist before γ'' can precipitate^[8,13]. There is general agreement, though, that γ' does precede γ'' .

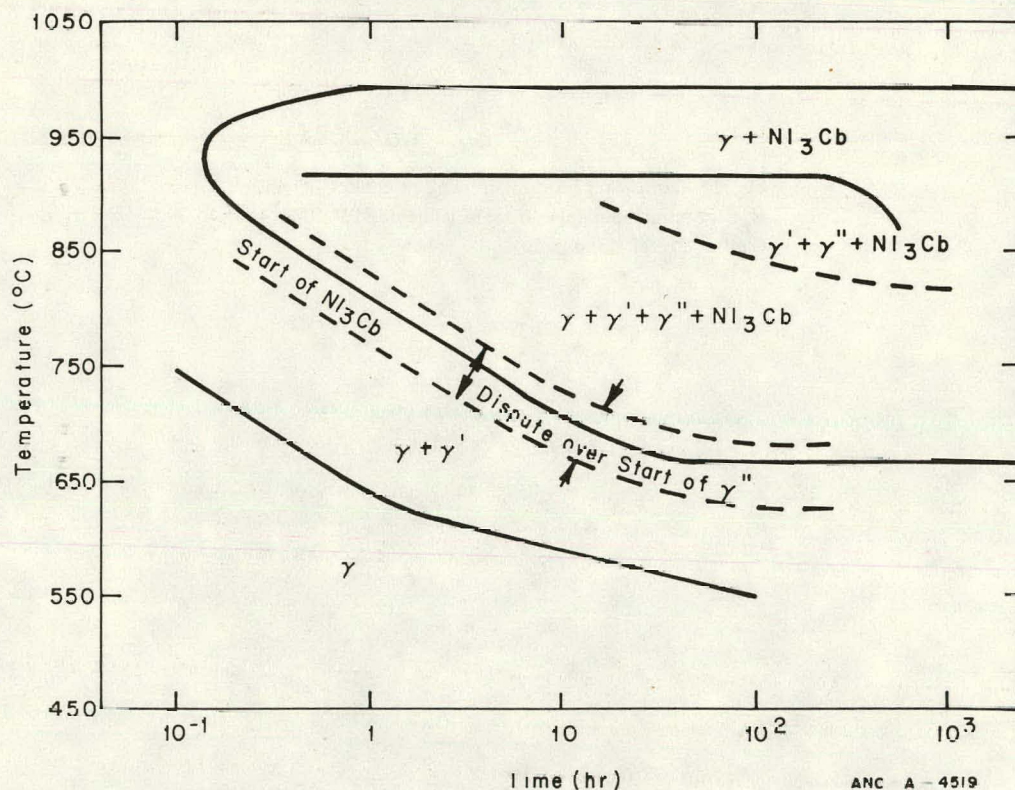


Fig. 17 Transformation diagram for Alloy 718^[17].

4.2 Mechanism of Overaging

Overaging is synonymous with loss of strength upon increase in temperature, which for Alloy 718 begins to occur at 1,200°F (649°C), especially in long time tests^[8]. Paulonis et al have concluded that overaging is due primarily to (a) coarsening of γ'' precipitates and (b) partial solution of γ'' and γ' with concurrent formation of stable

orthorhombic Ni_3Cb . They found that γ'' particles begin to coarsen very rapidly above $1,400^\circ\text{F}$ (760°C). The ripening (coarsening) γ'' particles appear to maintain their coherency up to $1,600^\circ\text{F}$ (871°C) where they begin to transform into Ni_3Cb needles. They observed that γ' does not ripen above $1,400^\circ\text{F}$ (760°C) in a manner similar to that observed for γ'' . This can be deduced from Table II, where test results of measurements of average particle sizes obtained from various heat treatments between $1,200$ and $1,600^\circ\text{F}$ (649 and 871°C) are tabulated^[8]. The results of measurements of particle sizes for Alloy 718 are compared with particle sizes obtained for similarly heat treated Waspaloy which is primarily hardened by γ' . With the use of optical metallography, they found that between $1,550$ and $1,660^\circ\text{F}$ (843 and 904°C), substantial amounts of stable orthorhombic Ni_3Cb phase formed. They suggested that much of the γ'' dissolved to provide columbium for the stable Ni_3Cb phase. Since stable Ni_3Cb contributes little to strength, weakening occurs when this phase forms. Based on chemical extraction, Paulonis et al^[8] found that samples heated to $1,600^\circ\text{F}$ (871°C) contained only 9 wt% precipitate ($\gamma'' + \text{stable Ni}_3\text{Cb}$), whereas fully heat treated material contained 19 wt% ($\gamma'' + \gamma'$).

4.3 Thermomechanical Effects

Cremisio et al^[18] found that the amount of stable Ni_3Cb phase precipitated at $1,750^\circ\text{F}$ (954°C) is dependent on the amount of straining prior to the $1,750^\circ\text{F}$ (954°C) heat treatment. They observed that strain causes the stable Ni_3Cb to precipitate in banded areas where segregation of columbium-containing particles (possible CbC) from solution has previously occurred. Concentration of stable Ni_3Cb in such areas increases with increased amounts of imposed strain^[18]. It was observed also^[18] that overaging of the hardening precipitates (γ'' and γ') was accelerated by straining at aging temperatures. This accelerated the overaging more than straining prior to exposure at aging temperatures without superimposed straining.

Barker et al^[19] have studied the effects of phase instabilities in Alloy 718 on creep rupture specimens. The specimen material from the creep rupture tests had been subjected to temperatures ranging from $1,000$ to $1,300^\circ\text{F}$ (538 to 704°C) and times to 33,000 hours. They concluded that there were no detrimental structural instabilities based on these temperatures and times, except for expected overaging at $1,300^\circ\text{F}$ (704°C). Some interesting observations were noted, however. Alpha prime (chromium-rich solid solution) and sigma were present upon long-time exposure at $1,200$ and $1,300^\circ\text{F}$ (649 and 704°C). Sigma phase is rarely detected in Alloy 718. They felt that possibly the reason for the occurrence of sigma phase was depletion of certain elements, during long-time testing, which may have changed the local composition enough to cause sigma phase to form. Sigma phase in substantial amounts is deleterious to ductility and notch toughness. They were also able to correlate changes in mechanical properties with formation of stable Ni_3Cb ; the stable Ni_3Cb formed possibly at the expense of γ' and γ'' . They also noted that there appeared to be a difference in the Ni_3Cb that formed at high temperatures [$1,700$ to $1,800^\circ\text{F}$ (923 to 982°C)] and that which forms at aging temperatures. There was a difference in chemical behavior in the extraction media between the two products. Finally, they noted that increasing the stress level accelerates the $\gamma' \rightarrow \gamma'' \rightarrow \text{Ni}_3\text{Cb}$ (orthorhombic) reaction.

4.4 Effect of Chemistry Variations

Chemistry variations can have an effect on structural stability in Alloy 718. Rizzo and Buzzanell^[11] studied chemistry variation effects on structural stability. They prepared

TABLE II

APPROXIMATE AVERAGE PRECIPITATE PARTICLE SIZES
IN ALLOY 718 AND WASPALOY

Heat Treatment	Alloy 718		Waspaloy
	γ' (diam of sphere - $\frac{\circ}{A}$)	γ'' (diam of disc - $\frac{\circ}{A}$)	γ'' (diam of sphere - $\frac{\circ}{A}$)
Fully heat treated (FET)	250	600	450
FHT + 1,200 ^o F (649 ^o C)/100 hours	300	725	-
FHT + 1,400 ^o F (760 ^o C)/100 hours	600	3,000	550
FHT + 1,550 ^o F (843 ^o C)/16 hours	750	3,500	750
FHT + 1,600 ^o F (871 ^o C)/16 hours	[a]	5,500	[a]

[a] Phase is completely dissolved.

heats of Alloy 718 with chemistry variations relative to nominal specifications of columbium, titanium, aluminum, and silicon. Of all elements studied, columbium variations appeared to have the greatest effect on material exposed to 1,200°F (649°C) for 500 hours. High strengths and low ductilities were noted with increased amounts; Ni₃Cb and Laves phases were also more prevalent. Low columbium content resulted in more gamma prime and CbC. Increased aluminum content favored larger amounts of Laves phase but had variable or little effect on mechanical properties. Low aluminum content appeared to increase the tendency for formation of stable Ni₃Cb. Higher strengths and increased amounts of gamma prime were noted with high titanium contents. As expected, low titanium content favored improved ductility. High silicon content promoted the formation of Laves phase and M₆C. Low levels of silicon favored increased formation of stable Ni₃Cb and CbC.

4.5 Grain Size Effects

Grain size effect on phase stability of Alloy 718 has been studied by Stroup and Heacox^[20]. Their study considered the effects of uniform grain size (ASTM 7 or finer), duplex grain size, and coarse grain size (ASTM 5 or coarser) on stress rupture properties before and after exposing specimen material at an applied stress of 50 ksi to 1,250 and 1,300°F (649 and 704°C) for 500 hours. For uniform grain size, the optimum grain size was reported to be ASTM 6 to 7 for the best balance of stress rupture properties, especially before exposure. A risk of notch brittleness becomes serious with a uniform grain size of ASTM 5 or coarser; a serious loss in strength occurs with a size smaller than ASTM 7. Apparently, duplexing can be tolerated if the coarse grains do not exceed 50 to 60% or exceed ASTM 4 in size. Microstructurally, they found that stress rupture properties are reduced if Ni₃Cb needles are large, which happens if the grains are large. However, stress rupture properties are improved if the volume of small Ni₃Cb needles is increased.

4.6 Technique to Improve Phase Stability

Alloy 718 begins to suffer significant strength reduction at 1,200°F (649°C). This is attributed to the coarsening or ripening of the major hardening constituent γ'' (DO₂₂ structure). Cozar and Pineau^[13] have pursued the possibility of reducing the ripening effect by varying the (Ti + Al)/Cb ratio to increase the γ' constituent which is known not to coarsen as fast as γ'' . They have found a "compact morphology" which proved not to coarsen very fast. It has a microstructure consisting of cube-shaped γ' particles coated with a γ'' shell over the six faces. Apparently, an improvement in thermal stability has been obtained by them with this structure in related 718 alloys. Alloy designers should be encouraged by this bit of information.

5. STRENGTHENING MECHANISM

The evidence^[5,8,9,13,21,22] appears overwhelmingly to favor the concept that γ'' is the major hardening constituent in Alloy 718 rather than the γ' . This concept has been developed mostly from results of electron diffraction studies.

The majority of the strengthening associated with the γ'' comes with coherency straining^[9] rather than antiphase boundary hardening associated with γ' . The coherency

straining results from atomic mismatch between the γ and γ'' . Oblak et al^[9], using developed models for both order strengthening and coherency strengthening, were able to show that the critical resolved shear stress for hardened Alloy 718, under various stressing situations, was more nearly approximated by the coherency strengthening mechanism when compared to actual experimental data than by the order strengthening method.

6. MODES OF DEFORMATION

Deformation in Alloy 718 occurs by massive planar slip in a heterogeneous fashion^[8]. Planar slip, instead of the more homogeneous wavy glide slip, is probably favored because nickel, being the major constituent, has a low stacking-fault energy. Solids with low stacking-fault energies do not commonly cross slip with screw dislocations. Therefore, dislocations will not naturally glide out of their slip planes to move around obstacles such as precipitated γ'' particles. This being the case, dislocations must "cut" through the γ'' particles instead of "bypassing". However, beyond a certain γ'' particle size, bypassing can occur by either looping or, if the temperature is high enough (usually $0.5T_m$), by dislocation climb. Particle size of sufficient magnitude for looping can occur during overaging when γ'' particles coarsen. At temperatures below 1,200°F (649°C), Alloy 718 will almost always exhibit particle cutting by dislocations because particle coarsening to promote dislocation bowing and high enough temperatures to promote dislocation climb are nonexistent.

Oblak et al^[9] reported from their studies that there is no difference in deformation mode between material normally aged and that aged under stress. They observed that shear takes place by dislocation pairs, i.e., by $a/2 \langle 110 \rangle$ pairs. They found no evidence for shear by a stacking-fault mode, thus substantiating their models and the observations of Paulonis^[8].

7. RECOVERY AND RECRYSTALLIZATION

Based on observations by the author and studies performed by Meyers^[23], recovery, recrystallization, and grain growth are thought to occur in the temperature ranges given in Table III. If the material is heavily cold worked (greater than 20% deformation), the low end of each range is applicable and, if lightly worked, the upper end of each region is applicable.

TABLE III

TEMPERATURE RANGES FOR RECOVERY, RECRYSTALLIZATION,
AND GRAIN GROWTH FOR ALLOY 718

<u>Condition</u>	<u>Temperature Range, °F (°C)</u>
Recovery	1,450 to 1,750 (788 to 954)
Recrystallization	1,650 to 1,950 (899 to 1,066)
Grain growth	1,800 (982) and above

8. HEAT TREATMENT EFFECTS

This section will be devoted to effects that heat treatment variables have on the microstructure and mechanical properties of Alloy 718. This evaluation will be centered around heat treatment variables associated with the two common commercial heat treatments for Alloy 718. These are described in Table IV.

TABLE IV

COMMON HEAT TREATMENTS GIVEN ALLOY 718

Heat Treatment 1:

Solution anneal at $1,750 \pm 25^{\circ}\text{F}$ ($954 \pm 14^{\circ}\text{C}$), hold for 1 hour, air cool. Age at $1,325 \pm 15^{\circ}\text{F}$ ($718 \pm 8^{\circ}\text{C}$), hold 8 hours, furnace cool at the rate of 100°F (55°C)/hour to $1,150^{\circ}\text{F}$ (622°C) and hold at $1,150 \pm 15^{\circ}\text{F}$ ($622 \pm 8^{\circ}\text{C}$) for 8 hours for a total aging time of 18 hours; air cool.

Heat Treatment 2:

Solution anneal at $1,925 \pm 25^{\circ}\text{F}$ ($1,052 \pm 14^{\circ}\text{C}$), hold for 1 hour, air cool. Age at $1,400 \pm 15^{\circ}\text{F}$ ($760 \pm 8^{\circ}\text{C}$), hold for 10 hours, furnace cool at the rate of 100°F (55°C)/hour to $1,200^{\circ}\text{F}$ (649°C) and hold at $1,200 \pm 15^{\circ}\text{F}$ ($649 \pm 8^{\circ}\text{C}$) for 8 hours for a total aging time of 20 hours; air cool.

8.1 Solution Annealing

Some very interesting work has been done by Muzyka and Maniar^[24] on determining the effects of solution-annealing treatments on the microstructure and properties of hot rolled 718 alloy. Their work in many ways corroborates the findings of Eiselstein^[2] on this matter. They found that room temperature tensile strength and stress rupture ductility

decrease with increasing solution-anneal temperature, such as from 1,750°F (954°C) (Heat Treatment 1, Table IV) to 1,925°F (1,052°C) (Heat Treatment 2 Table IV). Also, they reported that stress rupture life increased with the higher solution-anneal temperature (Heat Treatment 2). The effects of high and low temperature solution-anneal treatments on tensile and stress rupture properties are listed in Table V, which is data reported by Hall^[3]. Similar data presented by International Nickel Co. confirm these findings. The data in Table V show that one solution-annealing temperature does not necessarily give the optimum tensile and stress rupture properties simultaneously.

TABLE V

EFFECT OF ANNEALING TEMPERATURE ON THE ROOM TEMPERATURE TENSILE PROPERTIES AND STRESS RUPTURE PROPERTIES OF AGED ALLOY 718^[3]

Annealing Temperature, °F (°C)	R. T. Tensile Properties		Stress Rupture Properties	
	0.2% Yield Strength (ksi)	Elongation (%)	Life (hr)	Elongation (%)
1,750 (954)	178	7	17	29
1,900 (1,038)	166	24	152	7

Heat Treatment:

1 hour annealing temperature, air cool
 8 hours 1,325 °F (718°C), furnace cool to 1,150°F (621°C), air cool

Stress Rupture Conditions:

1,300°F (704°C), 75 ksi,
 no notch failures

Composition:

0.75% Al and 1.45% Ti (neither ideal)

The studies by Muzyka and Maniar^[24] also showed that sample material solution treated at 1,850 or 1,900°F (1,010 or 1,038°C) and fully aged exhibited relatively clean grain boundaries with occasional Cb(C,N). Some M₆C at grain boundaries was found in material solution-annealed at 1,850°F (1,010°C) but not with anneals above that temperature. Presumably, M₆C could be found at grain boundaries in material annealed below 1,850°F (1,010°C). They^[24] also found that grain growth does not occur in hot rolled 718 alloy bars until a temperature of about 1,850°F (1,010°C) is reached. Ni₃Cb, they felt, controls grain size for solution treatments below 1,800°F (982°C), and M₆C and CbC control for temperatures above 1,800°F (982°C).

8.2 Aging Treatment

A single aging temperature is not commonly used on Alloy 718. When it is, maximum strength can be attained by heating between 1,300 and 1,400°F (704 and 760°C) for 10 to 20 hours^[2,3]. Lower temperatures than these require excessive times. More commonly, as noted in Table IV, two aging temperatures are used, where the second temperature is about 200°F (110°C) below the first. This method increases strength by 10 to 20 ksi^[3] over a treatment with a single aging temperature. It is interesting, though, that a slow cool, 20°F (11°C)/hr, from the first aging temperature of double age will provide the same strength results^[3]. One possible explanation for the increase in strength of the double aging versus the single aging treatment is that the matrix material at the lower temperature of a double age experiences an equilibrium change and cannot keep as much hardening precipitate in solution and, therefore, allows more to be precipitated.

8.3 Anticipated Microstructures

Sufficient information has been generated at this point so that one can predict the microstructure that will result upon subjecting Alloy 718 to the various heat treatments commonly made on this material. Consider, for example, Heat Treatment 1 of Table IV. The conditions of this treatment are superimposed in Figure 18 on the TTT diagram for Alloy 718. The figure shows that during the 1,750°F (954°C) anneal, stable Ni₃Cb is precipitated. This is the high temperature Ni₃Cb that Barker et al^[19] mentioned. Further, on aging at 1,350°F (732°C) for eight hours, additional amounts of the stable Ni₃Cb will form along with the hardening constituents γ' and γ'' . The Ni₃Cb formed at 1,350°F (732°C) will have a different chemistry from that formed at 1,750°F (954°C)^[19]. Finally, only $\gamma' + \gamma''$ will form during the second portion [1,150°F (621°C)] of the double aging cycle. The final product will consist of γ' and γ'' uniformly distributed in the matrix (γ) and Ni₃Cb at grain boundaries. It is interesting to note that if an 1,800°F (982°C) anneal temperature (upper reaches of Heat Treatment 1, Table IV) had been used for this example rather than the 1,750°F (954°C) anneal, Laves phases would have been identified as a possible constituent forming during the solution-anneal stage of the heat treat cycle. It is probably for this reason that few people specify 1,800°F (982°C) in their heat treat schedules for this alloy.

Now consider Heat Treatment 2: 1,950°F (1,066°C) solution anneal plus 1,400/1,200°F (760/649°C) age, described in Table IV. The conditions of the heat treatment are approximately given in Figure 19. Since the 1,950°F (1,066°C) anneal is above the solvus lines for both Ni₃Cb and Laves phase, neither of these will form during the solution anneal, and neither will form during cooling to 1,400°F (760°C), the start of the aging cycle. Therefore, the only part of the heat treat cycle shown in Figure 19 is the aging portion. Clearly, the diagram shows that during the 1,400°F (760°C) aging, hardening constituents (γ' and γ''), Ni₃Cb, and CbC film form during the ten hours at this temperature. A CbC film forms during this heat treatment because it dissolves during the 1,950°F (1,066°C) anneal and, therefore, will reprecipitate during the aging cycle^[2]. [This is not the case during the 1,750°F (954°C) solution anneal; therefore, the formation of a CbC film is not shown in Figure 18 for the 1,750°F (954°C) anneal condition.] The volume of the CbC film in the grain boundaries will probably be greater following the age cycle at 1,400°F (760°C) than prior to anneal since some grain growth will probably occur during the 1,950°F (1,066°C) anneal. Further aging at 1,200°F (649°C) will produce only γ' and γ'' to increase strength. The final product will be similar to that produced with the 1,750°F

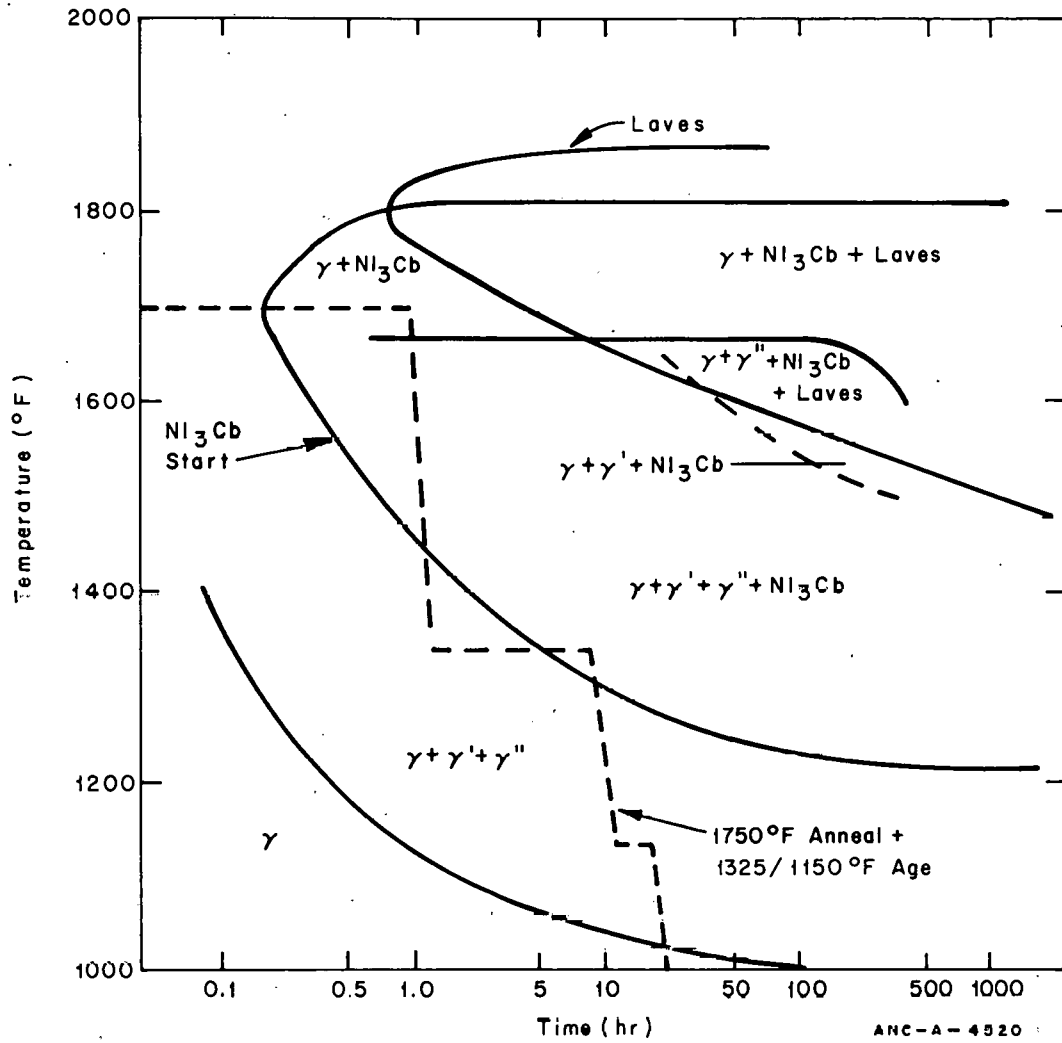


Fig. 18 Transformation diagram for Alloy 718 showing commercial heat treatment of 1,750°F (954°C) anneal + 1,325/1,150°F (718/621°C) age.

(954°C) solution anneal plus 1,325/1,150°F (718/621°C) except a noticeable NbC film will exist in the grain boundaries along with other possible carbides.

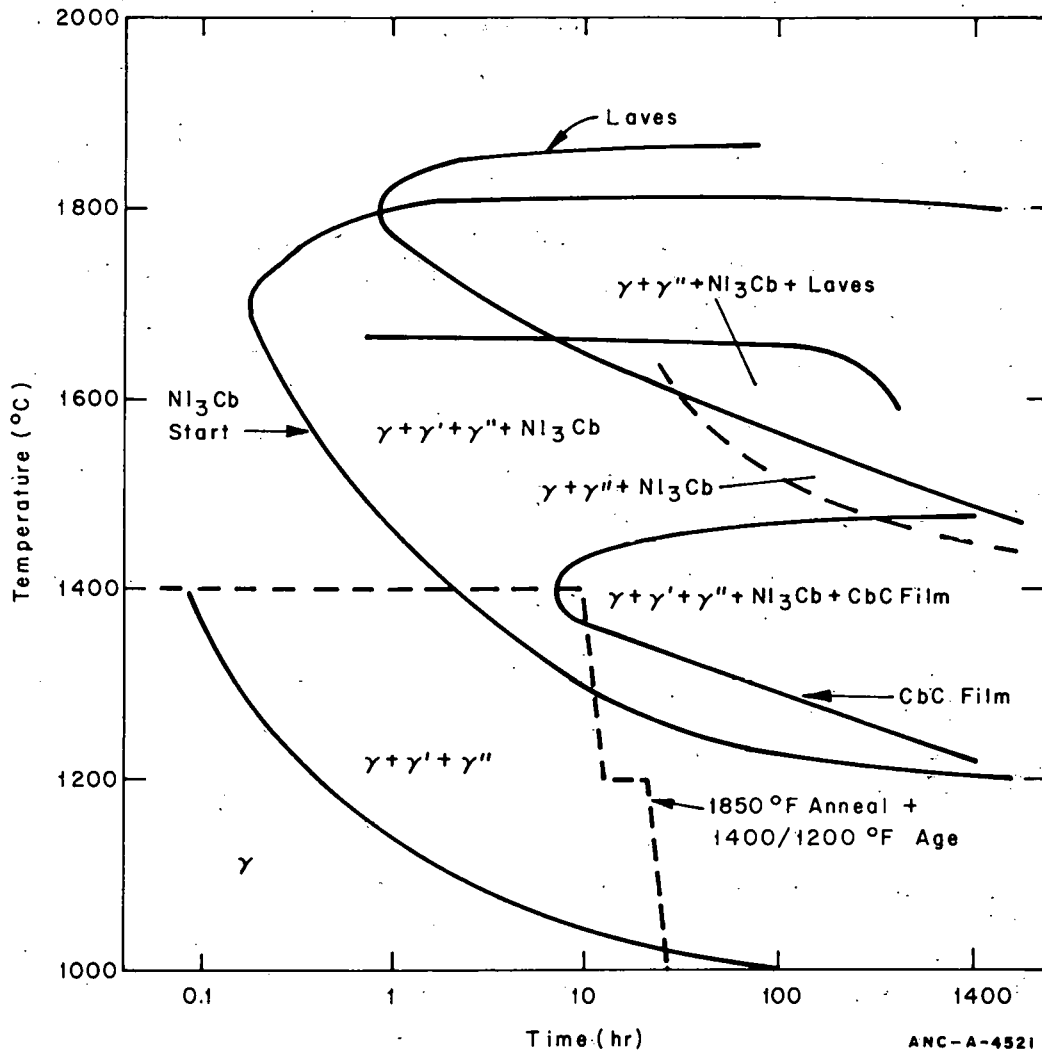


Fig. 19 Transformation diagram for Alloy 718 showing commercial heat treatment of 1,950°F (1,066°C) anneal + 1,400/1,200°F (760/649°C) age.

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