ANALYSIS AND ELIMINATION OF TIME DEPENDENT NOTCH SENSITIVITY IN ALLOY 718

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ABSTRACT

The Mechanism for High Temperature Time Dependent Notch Sensitivity of alloy 718 is characterized by combining the dislocation theory of precipitation hardening with plane stress experimental results at elevated temperatures. The criteria for elimination of Time Dependent Notch Sensitivity is formulated by implementation of a heat treat procedure that increases the average distance between precipitates in alloy 718 from 50 to 167 atomic spaces, while maintaining the minimum required yield stress. Here, the flexible dislocation passes between precipitate particles without cracking them. This heat treat considerably reduces the detrimental orthorhombic delta phase (allotropic transformation of the body centered tetragonal gamma double prime precipitate phase) that has formed in service. It is shown by analysis, that the critical precipitate size decreases with decreasing solution temperatures, (with solution time and aging treatment held constant), but the length of dislocation line increases to the critical length, and is the limiting factor that gives an alloy its maximum yield strength. The precipitate size is however, very important when the particle is sheared by a dislocation, since it is the precipitate size and energy to break its atomic bonds that determines if the dislocation will crack the precipitate or circumvent it. Finally, it is shown by analysis and experiment that hardness testing is inadequate to evaluate a production heat treat precipitation hardening procedure and should be replaced by a shear or tensile yield point measurement, combined with an analysis of the atomic dispersion size. This work is one of the first applications of dislocation theory to directly solve a reliability problem in production.

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A-NOTCH SENSITIVE CRACKING AND PRECIPITATE DEPLETION

A. INTRODUCTION.

The in-service degradation of alloy 718 components parts is a two fold problem consisting of:

a. The mechanism responsible for premature crack initiation and

b. A solid state transformation that depletes the material of its hardening precipitates, thereby weakening the material. The first of these phenomena called Time Dependent Notch Sensitivity is known to occur in two alloys within the Air Force inventory: Waspaloy and alloy 718. Time dependent Notch Sensitivity is the susceptibility for premature crack initiation to occur at elevated temperatures, in the vicinity of a notch or stress raiser, as a function of time at a constant external stress below the yield stress as compared to an identical unnotched specimen. This is followed by subsequent crack propagation by any number of mechanisms to critical crack length and ultimate fracture. Time Dependent Notch Sensitivity is a phenomena where premature crack initiation occurs at elevated temperatures, caused by the effects of ordinary room temperature plastic deformation combined with plastic deformation due to creep. The second of the detrimental effects is the transformation of γ''(gamma double prime), a coherent hardening precipitate of chemical compound (Ni_3Cb). This precipitate exists in a metastable body centered tetragonal structure and is allotropically transformed into an orthorhombic noncoherent δ (delta) phase of the same chemical composition:

$$\underset{\text{Body Centered Tetragonal}}{\gamma''\ (Ni_3Cb)} \longrightarrow \underset{\text{Orthorhombic}}{\delta\ (Ni_3Cb)} \qquad (1)$$

This transformation depletes the alloy of its strengthening properties and depending on the form taken by the resulting delta phase, it may impart stress concentration inside the grains.

B. Experimental Results.

Replicas normal to the fracture surfaces were made on service induced cracks. The replications were examined under a optical microscope and exhibited noncontinuous crack profiles indicating multiple initiation sites. This suggested that subsurface multiple origin microcracking may be occurring, with all microcracks not reaching the surface at the same time. Cross sections were mounted, polished, etched and studied. The observed microstructures exhibited a relatively high density of delta phase <u>platelets</u>, (Figure 1), formed due to the solid state transformation. The delta phase in alloy 718 is normally in a <u>globular</u> shape and it manifests itself primarily at the grain boundaries. This normal microstructure is shown in Figure 2. <u>Globular</u> delta phases impart certain desirable properties to the alloy such as pinning down of the grains for optimization of stress rupture ductility. However, the formation of <u>platelets</u> of delta, degrades the alloy by depleting the hardening phase γ''. The brittle delta platelets are an incoherent, orthorhombic, nickel and niobium rich phase that precipitates at elevated temperatures in a Widmanstätten-type structure. The platelets act as stress raisers throughout the material to detrimentally affect the high temperature fracture properties of the alloy. It is this transformation that sets the temperature time limits for engine applications. A cross section of an alloy 718 crack (Figure 3,4),

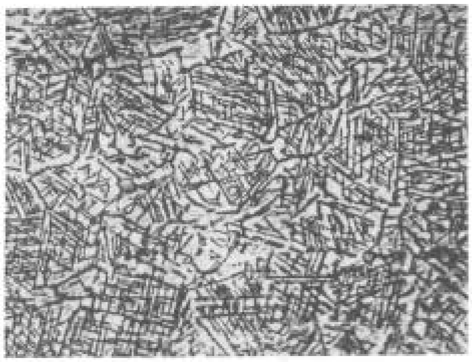

Figure 1 - (1000x) (Marbles etch)

Optical micrograph of microstructure of alloy 718, taken from a production item removed from service. The needles are the projection of the platelets of delta phase in a Widmanstatten structure.

Heat treat procedure:

Solution Treat - - -	1,700°F for 1 hour
Age - - - - - - - -	1,300°F for 8 hours
	1,150°F for 8 hours

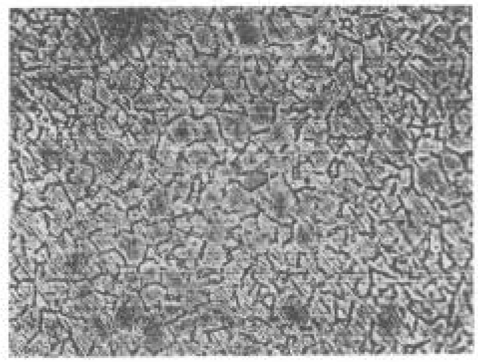

Figure 2 - (1000x) (Marbles etch)

Optical micrograph of structure of alloy 718 virgin material, taken from newly manufactured item that did not see service. Note the undeformed grains with globular delta phase at the grain boundaries.

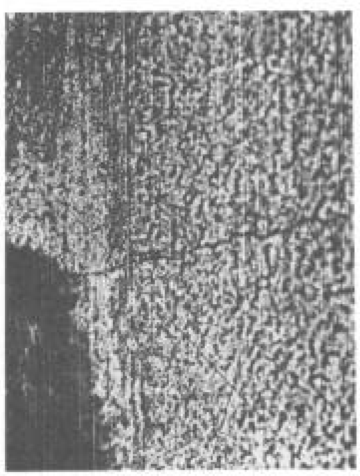

Figure 3 - (1000x) (Marbles etch)

Optical micrograph of a notch induced line crack profile in a failed Pratt and Whitney part.

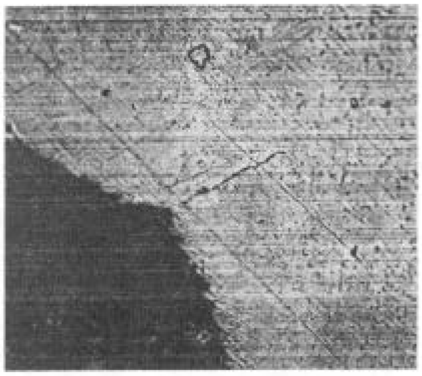

FIGURE 4 - (500x)

(Polished but not etched)
Optical micrograph of a notch induced intergranular microcrack, in the early stages of crack development. The part is a Pratt and Whitney engine component, manufactured from alloy 718 and heat treated with a notch sensitive procedure. (Service condition 1000°F-1200°F).

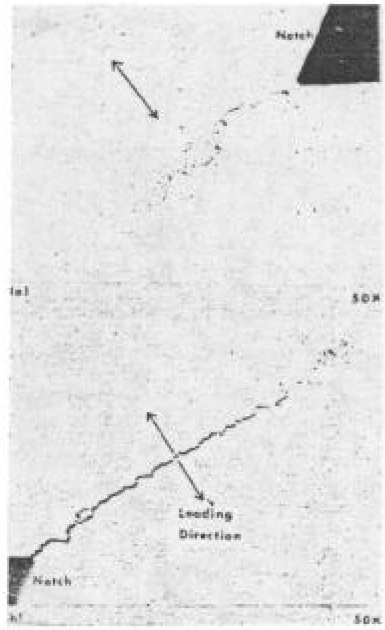

FIGURE 5 - (50x)

(Polished but not etched)
Optical micrograph of a double notched specimen of Waspaloy, heat treated with a notch sensitive procedure, tested at 1000°F at 80 ksi for 90% of its rupture life.

a. Microcracks are in an early stage of crack initiation.

b. "Through thickness" crack at a later stage of intergranular crack growth (Reference 1).

showing failure of the fracture surfaces to separate, displays the characteristics of a tight line crack rather than the usual wedge shape typical of other alloys. The resulting lack of occurrence of capillary action, combined with surface disturbances and disadvantageous orientation, obscures the crack beyond the detection limits of fluorescent penetrant and radiography techniques.

Consider a failed notched tensile specimen tested under a constant load below the yield point at a temperature between 1000°F and 1400°F. Examination of the fracture shows it to consist of two distinct portions. The initial part or slow growth portion began at the notch and extended intergranularly in a direction perpendicular to the loading direction and was discolored from oxidation. (Figure 5). The remainder or fast fracture portion (not shown), was a transgranular "through thickness" shear failure that was not discolored by oxidation(1).

Critical to the crack initiation, was localized plastic yielding at the base of the notch upon loading, followed by subsequent creep deformation at the grain boundaries(1). Both phenomena contributed to crack initiation, as they alleviated the stress concentrations introduced by the existence of a notch.(1) Slow crack extension then occurred primarily by plastic deformation, as evidenced by a coallescence of voids or dimpled region on the fracture surface, as the crack tip advanced across the specimen(1). Ultimate transgranular fracture occurred when the critical crack length corresponding to the external load was reached. The notch sensitive cracks initiated after 60% of the rupture life as compared to 95% - 97% in the unnotched specimen, where the intergranular cracks initiated randomly within the gauge length(1).

C. THEORETICAL DEVELOPMENT.

In order to develop the concept of crack initiation due to Time Dependent Notch Sensitivity, it is necessary to interpret its mechanism in terms of an analysis of precipitation hardening of coherent dispersed particles. The theory is attributed to Mott, Nebarro, Cottrell and Orowan(2). A complete theory of precipitation hardening must cover the range of values the flow stress takes from the solution treated condition before the saturated solid solution has decomposed, (zero size precipitates) up until the overaged condition where the precipitates are so large that they fail to adequately act as a strengthening mechanism.

To strengthen a material the increase in yield stress depends largely on the strength, structure, spacing, size, shape, and distribution of the precipitate particles as well as on their degree of misfit. When an isolated solute atom is present in the solution treated condition, it creates a degree of misfit ϵ in the solvent crystal, so that the atomic radii of the solvent and solute atoms are given by r_a and $r_a(1+\epsilon)$ respectively. Any size coherent spherical particle that encloses a given group of lattice sites, changes its radius by the factor $(1+\epsilon)$ so that the degree of misfit of a spherical particle is independent of its size. The precipitate particles are dispersed in the alloy material to create a strain field that produces residual microstresses that act as barriers to dislocation motion and gives the material its yield strength. The shear strain associated with these microstresses at a distance R from a spherical inclusion of radius r_o and misfit ϵ is given by the theory of elasticity to be:

$$\text{Strain} = \frac{\epsilon r_o^3}{R^3} \qquad (R > r_o) \tag{1}$$

Since ϵ is constant, regardless of precipitate size, softening occurs when changes in the state of dispersion take place, without changes in ϵ. Lets assume that the yield point stress of an alloy is due to the average arithmetic mean of the internal stress that acts at a distance from a given dislocation line segment, due to coherent spherical precipitate particles dispersed throughout the material matrix. Mott and Nebarro(2) derived* an expression for this average internal shear stress or yield point stress that immobilities a dislocation:

$$\sigma_i = 2\, G\, \epsilon\, C \tag{2}$$

Where:

G = Shear modules of elasticity $\frac{\text{lbs}}{\text{in}^2}$

ϵ = Degree of misfit $\frac{r_o - r_i}{r_i}$ $\frac{\text{in}}{\text{in}}$

r_o = Radius of solute in

r_i = Radius of solvent in

C = Volume fraction of precipitate volume/volume

Note that this result for σ_i the flow stress, is independent of: (a) Ω, the distance between precipitates, (b) the crystallographic structure of the alloy and (c) the dislocation density, but depends very strongly on: (a) the shear modulus of the matrix material, (b) the degree of misfit of the precipitate in the host material and (c) the volume fraction of the precipitate when the alloy is at its maximum yield point.

If one now considers a rigid straight dislocation line segment, equation(2) suggests that the solute atoms should fully harden the alloy, regardless of dispersion size, since it is independent of Ω. Alternatively, one can say that a straight dislocation line is acted on by randomly alternating stress fields, some of which push the individual dislocation segments forward and others that push other segments backward, so that the algebraic average of all stresses cancel one another and there is no hardening. Mott and Nebarro(2) then postulated that the theory was incomplete, since the dislocation line was not rigid, and individually pinned segments have the flexibility to move independently of neighboring segments to bend round regions of high interaction energy so the random forces do not all cancel. This would explain the effect of the scale of dispersion. Hence, the extent to which one section of the line can move independently of its neighbors, depends on the distance between them, Ω. Cottrell(2) then related the limiting radius of curvature R, to which a pinned dislocation line segment can be bent to the flow stress σ_i, by the principle that a curved dislocation line can be held in equilibrium in

* Let the average distance from any point in the matrix to the nearest precipitate particle be $1/2(N)^3$ in a material containing N particles per unit volume, each of radius r_o and misfit ϵ. Substituting in eq(1) gives a strain of approximately $8\epsilon r_o^3 N$, and when the concentration of solute $C \simeq (4/3)\pi r_o^3 N$, is inserted, equation(2) is obtained.

the shape of a curve, only when it is acted on by an outward shear force due to an applied external stress[2, 3]. The shear force acts normal to the dislocation line segment and is balanced in equilibrium, by an opposing inward force component due to the line tension at the ends of the element. Balancing these two forces, he arrived at the following result[2, 3]:

$$R = \frac{\alpha G b}{\sigma_1} \qquad (3)*$$

with α having the value of .5.
G = Shear modules of elasticity
b = Burgers Vector
σ_1 = Flow stress

Cottrell[2] then compared R with Ω, and showed that the maximum radius of curvature that a dislocation line segment can be bent corresponds to the maximum yield stress of the fully aged alloy when R is of the order of magnitude of Ω_c:

$$\Omega_c \simeq R = \frac{.5\, G\, b}{\sigma_1} \qquad (4)$$

When $\Omega = \Omega_c$ the dislocation can bend the maximum amount possible round regions of high interaction energy, and when slip occurs, the line segment overcomes the maximum potential energy barrier from one energy valley to the next.

Since σ_1 is independent of Ω in equation(2), Cottrell substituted(2) into(4) and used his values of ϵ = 1/5 and C = 1/40 to establish his criteria for the critical dispersion size[2].

$$\Omega_c \simeq R = \frac{.5\, G\, b}{\sigma_1} = \frac{.5b}{2\epsilon C} = \frac{.5b}{(2)(1/5)(1/40)} = 50b \qquad (5)$$

For the alloy to be in the fully aged condition, the average distance between precipitates is in the order of 50 atomic distances. When $\Omega < \Omega_c$ the material is underaged and is softened. When $\Omega > \Omega_c$ the material is overaged and is softened, as the yield point decreases with increased aging time, (Figure 6). This suggests that the precipitate size increases as the distance between precipitates exceeds the limiting radius of curvature, for the critical dispersion size given in equation(4) by Mott, Nebarro and Cottrell[2]. Orowan[2] later proposed, that in an overaged alloy, the dislocation line segments are pinned by widely spaced obstacles and do not have to overcome a large potential barrier to initiate the flow stress, but bulge instead into the spaces between obstacles. As this occurs the applied stress increases, until adjacent sides of neighboring loops join together to bypass the obstacles, so that the line segments pass between them. The precipitates are left encircled by small ring segments of the released dislocation, which act as additional obstacles that increase the flow stress on subsequent dislocations. The flow stress required for yield to occur is the shear stress that will bend the dislocation line segment into loops of $\Omega/2$. By equating the limiting of radius of curvature $\frac{\alpha G b}{\sigma_1}$ to $\Omega/2$, Orowan arrived at:

$$\sigma_1 = \frac{2\alpha G b}{\Omega} \qquad (6)$$

Where α = .5.

* Equation 3 is derived in Ref(3)

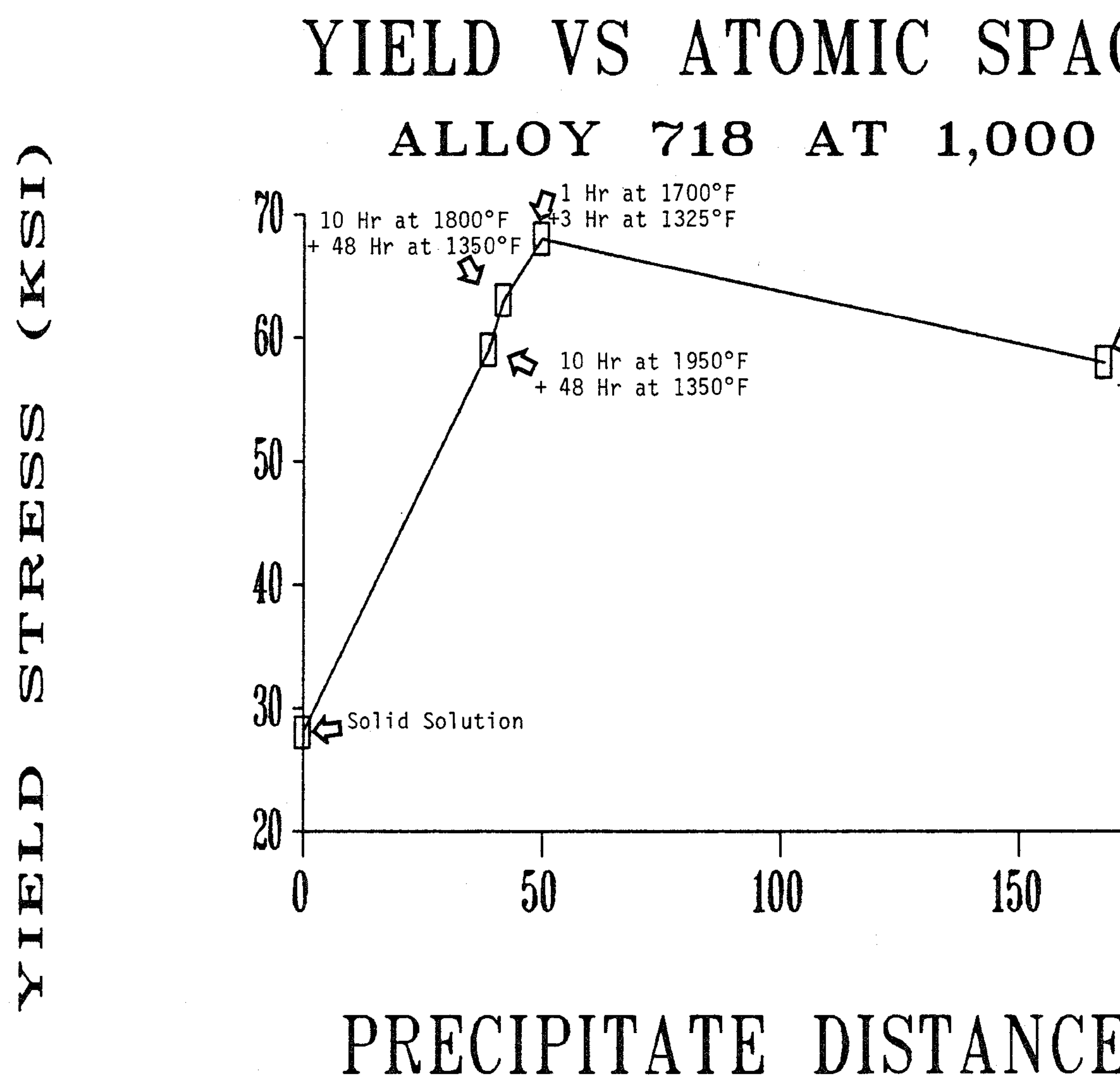

Figure 6a - Yield point as a function of atomic distance between precipitates or aging

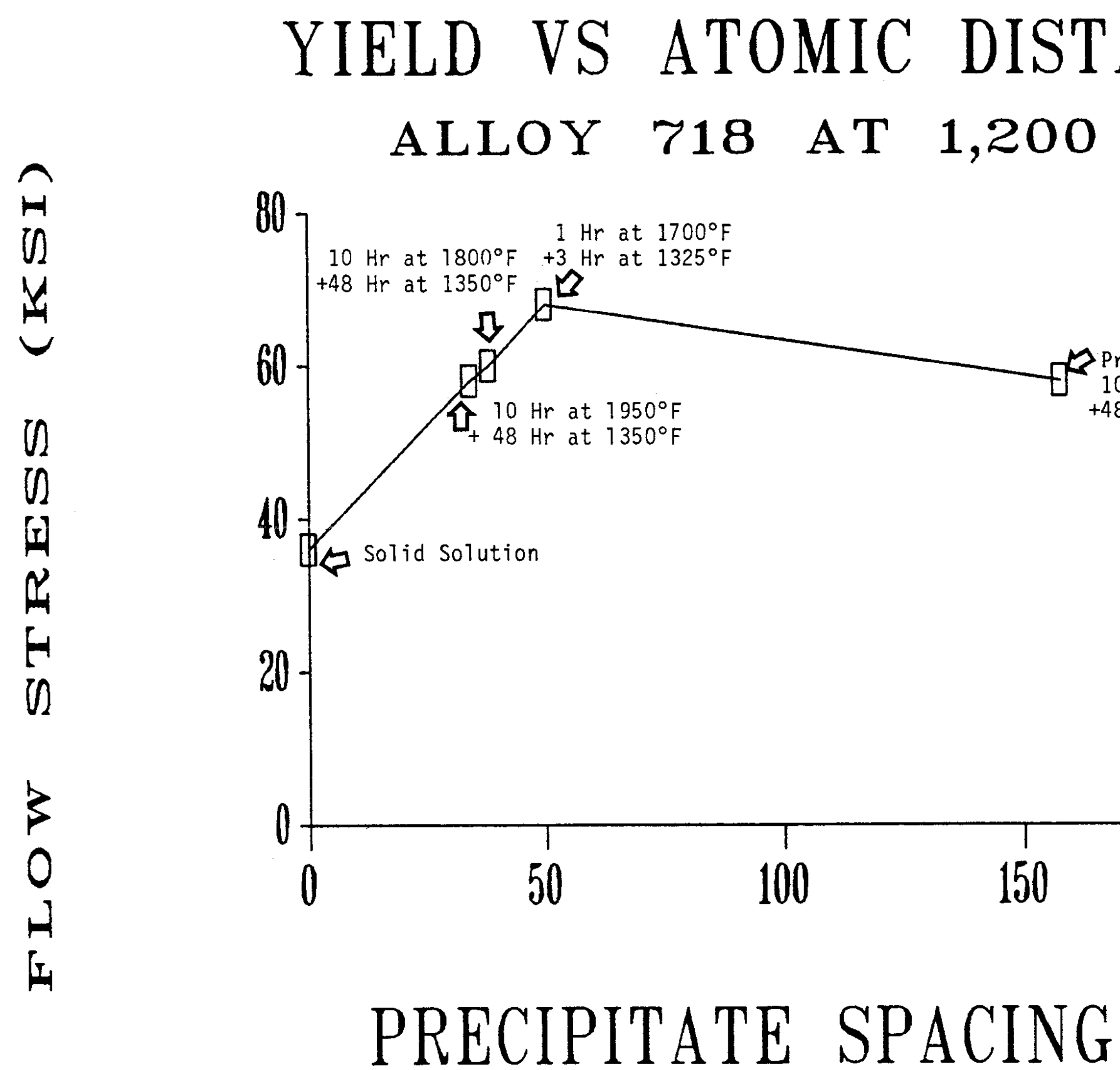

Figure 6b - Yield point as a function of atomic distance between precipitates or ag

Hence, the flow stress decreases inversely with the scale of dispersion for the overaged alloy. As precipitate growth takes place, Orowans above process replaces that of Mott, Nebarro and Cottrell(2) and the dislocations pass between the precipitates, <u>almost as soon as the scale of dispersion exceeds the critical value</u>(3, 4, 5). This precludes the premature shearing of precipitates by the dislocations.

Consider now the early stages of aging from the complete solution treated condition, up to the fully aged alloy. Here, the Orowan condition does not hold, since the line segment must reach the maximum radius of curvature and first satisfy equation(5) before it can pass between particles according to equation 6. Suppose instead that the dislocation moves without appreciably departing from a straight line. Let τ be the energy per unit area of the interface produced when the particle is sheared, and πr^2 be the area of a precipitate of radius r. Then the energy to cut the particle is given by $\tau(\pi r^2)$.

If b is the Burgers vector of the dislocation line segment of length Ω between precipitates, and the shear force per unit length acting along the dislocation line is $\sigma_{ic}b$. Equating the energy required to the work done by a dislocation as it moves:

$$\sigma_{ic} b \,\Omega\, 2r = \pi\, r^2 \tau \tag{7}$$

$$\sigma_{ic} = \frac{\pi}{2}\,\frac{\tau r}{\Omega b} \tag{8}$$

Where σ_{ic} is the flow stress necessary to shear the particle. For a coherent particle:

$$\text{if } \sigma_{ic} > \sigma_i \tag{9}$$

the dislocation will circumvent the precipitate. If:

$$\sigma_{ic} < \sigma_i \tag{10}$$

the particle will be sheared(4).

C-2 APPLICATION TO 718 ALLOY

Taking an actual case of alloy 718 alloy tested at 1000°F, the manufacturer specified the fully aged condition to have a tensile yield point of 135 ksi with the shear modules, $G = 9.7(10)^6$ psi at this temperature. Experimental results(6) related the shear yield strength to coherent γ'' and γ' precipitate particles 60A in diameter, after solution treating for 1 hour at 1700°F followed by ageing for 3 hours at 1325°F.

Predicting the flow stress theoretically, with the equation of Mott and Nebarro, (Equation 2) using the values of $\epsilon = 1/5$ and $C = 1/40$ given by Cottrell(2):

$$\sigma_i = 2\, G\, \epsilon\, C = 2(9.7)10^6\ \text{lb/in}^2\ (1/5)(1/40) = 97{,}000\ \text{psi} \tag{2}$$

This is the yield point stress for the fully aged alloy that immobilizes the dislocations, and that must be overcome if they are to circumvent their barriers. Substituting σ_i =96,000 PSI and G = 9.7(10)^6 psi into Cottrells equation gives the critical distance between precipitates:

$$\Omega = \Omega_c \simeq R = \frac{.5\ G\ b}{\sigma_i} \qquad (4)$$

$$\Omega_c = \frac{(.5)(9.7)10^6 b}{(97,000)} = 50b$$

This is in exact agreement with Cottrells prediction, presented earlier in equation (5) for the critical dispersion size Ω_c. The experimental results, measured the flow stress to be 67,500 psi. This was far below the value predicted here from Mott and Nebarros theory. This long range resisting force of 97,000 psi immobilized the dislocation, but since the precipitates were sheared at a lower stress:

$$\sigma_{ic} < \sigma_i \qquad (10)$$

The dislocation passed them by the cutting process. This is confirmed by transmission electron microscopy micrographs, taken after the specimen was creep rupture tested below the yield stress at 130 ksi, which showed that the deformation was localized in slip bands. This microstructure showed severe Time Dependent Notch Sensitivity, with ultimate fracture occurring after 17.5 hours in the notched specimen as compared to 5613.4 hrs for the unnotched specimen.

To find the energy necessary for the dislocation to shear the atomic bonds of the precipitate, when the flow stress, σ_{ic} = 67, 500 psi, is reached, equation (8) is solved for γ with:

Ω = 50b; r = 30 A

b = Burgers Vector for γ matrix = 3.59 A

then: $$\gamma = \frac{\sigma_{ic}\ 2\ \Omega\ b}{\pi\ r} = \qquad (8)$$

$$\frac{(67,500)\ \text{psi} \times (.06895)\ \dfrac{10^6\ \text{dynes}\ \text{cm}^3}{\text{psi}} \times (2.0)(50)(3.59(10)^{-8})^2\ \text{cm}^2}{(3.1416)\ (30)10^{-8}\ \text{cm}}$$

$$\gamma = 636.5\ \frac{\text{ergs}}{\text{cm}^2}$$

This is the energy required to cut through a γ'' particle of Ni_3Cb in the body centered tetragonal structure. This value is a reasonable result when compared to values exceeding 1000 ergs/cm^2 for intermetallic compounds incoherent with the matrix of aluminum or copper alloys, and 100 ergs/cm^2 for G.P. zones, coherent with aluminum base alloys(4).

The above analysis has shown that the limiting factor that determines the maximum yield strength of an alloy is the length of dislocation line that can be bent to the maximum radius of curvature, by the applied shear stress, and not the size of precipitate. The size of precipitate is however, very important when it is sheared by the dislocation, before the maximum radius of curvature is reached.

D. THE PROPOSED HEAT TREATMENT

In order to prevent Time Dependent Notch Sensitivity from occurring in alloy 718, the average atomic distance between precipitates must be increased to exceed the critical distance set forth by the theories of Mott, Nebarro and Cottrell(2). Then the dislocation line segments will bulge between obstacles, at a reduced flow stress, without shearing the particles. Of the different heat treat procedures formulated, all met the above requirements of Orowans criteria (equation 6), and all of them were free from Time Dependent Notch Sensitivity, when tested in plane stress at 1000°F and 1200°F(6). However, only one heat treatment* met the requirements formulated in AMS 5596 for yield and ultimate stress in plane stress deformation. Analyzing this heat treat, the results showed a flow stress of σ_i = 58 ksi measured at 1000°F (6) (TEM measurements showed a particle diameter of 200A)(6). (G = 9.7 10^6 psi at 1000°F).

Inserting this result to Orowans relation for an overaged alloy (equation (6)), gives the average dispersion size at 1000°F.

$$\Omega = \frac{2\alpha G b}{\sigma_i} = \frac{9.7(10)^6}{(58)\ 10^3} b = 167b$$

Thus, by applying this heat treat, the distance between precipitates has been increased from 50 to 167 atomic distances.

When the above flow stress, σ_i = 58 ksi required to overcome the yield stress of the material, is compared to the shear stress σ_{ic} necessary to cut through an Ni_3Cb precipitate particle, with:

$$\gamma = 636.5\ \underline{\text{ergs}}/\text{cm}^2\ ;\ \Omega = 167b$$
$$r = 100\text{Å};\ b \text{ for } \gamma \text{ matrix} = 3.59\text{Å}$$

$$\sigma_{ic} = \frac{\pi}{2}\frac{\gamma r}{\Omega b} = 134 \text{ ksi} \qquad (8)$$

The results satisfy inequality (9):

$$\sigma_{ic} > \sigma_i \quad \text{i.e.} \quad 134 \text{ ksi} > 58 \text{ ksi}$$

so that the dislocation will pass between the precipitates without shearing through them, and Time Dependent Notch Sensitivity does not occur.

This is confirmed by the TEM micrographs that showed that the dislocation were not localized in slip bands, nor observed cutting through the particles, but were observed entangled with and leaving loops around γ'' and γ' precipitates, after passing between them on the slip plane.

Comparing the rupture lives of notched specimens of the proposed heat treatment, with notched specimens of a heat treatment that exhibited the optimum tensile yield point shows a substantial increase in service life till fracture when Time Dependent Notch Sensitivity is eliminated (Figure 7).

* Solution treat 10 hours at 1700°F ±25°F, furnace cool with argon; age 48 hours at 1350°F ±15°F, furnace cool.

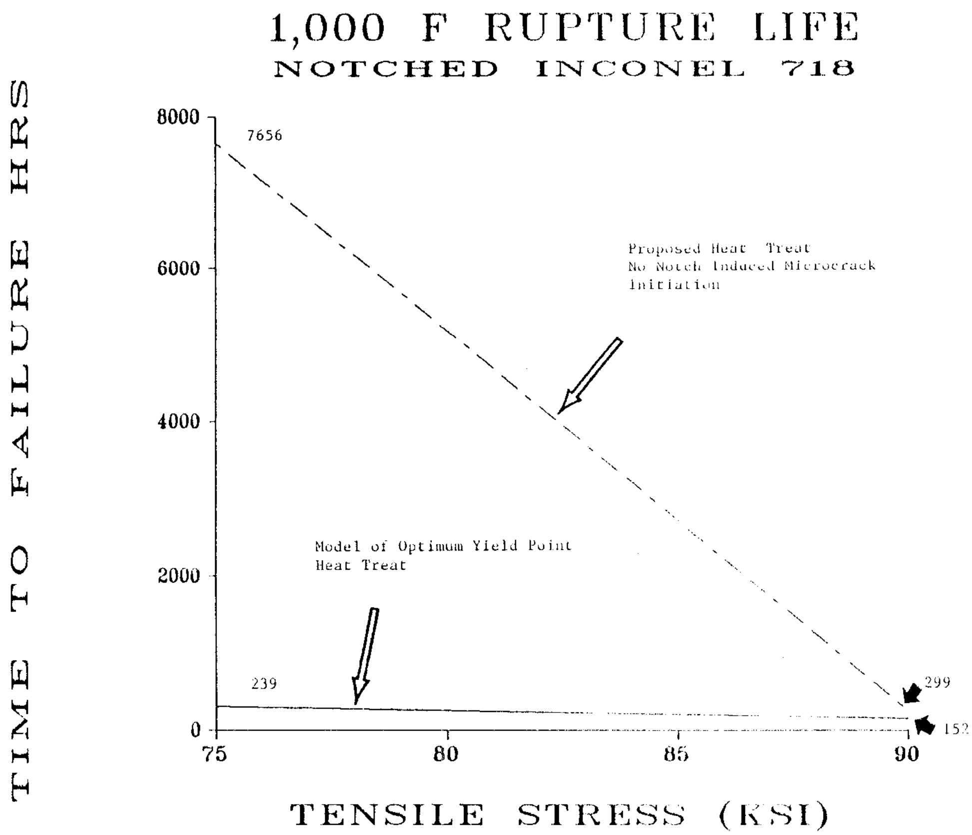

FIGURE 7: Comparison of time to fracture of existing optimum heat treat to proposed heat treat.

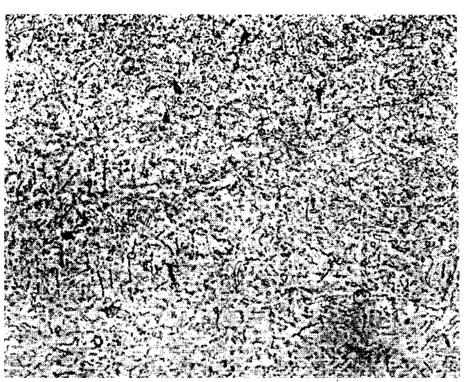

FIGURE 8 - (1000x) (Marbles etch)

Optical micrograph of the microstructure of alloy 718 taken from a part removed from service and given the proposed heat treat. The majority of the needles have been eliminated. Compare with the Windmanstatten needle structure of figure 1.

Of all the calculations performed on test data from different solution and aging heat treatments, the underaged and critically aged specimens had their precipitates sheared by dislocations. This resulted in premature crack initiation at elevated temperatures, (where subsequent creep grain boundary movement occurred), as compared with unnotched specimens. Initiation was followed by crack propagation to premature ultimate fracture. In each case TEM analysis exhibited slipbands due to localized non-homogeneous deformation[6], where the dislocations sheared the γ'' and γ' precipitates.

In the proposed treatment, the specimens were overaged and the dislocations passed between their precipitated barriers. Premature crack initiation at elevated temperatures did not occur as compared with unnotched specimens, and this resulted in extended service life (Figure 7).

TEM micrographs confirmed in each case, that the dislocations bypassed the precipitates without shearing them[6] resulting in homogeneous high temperature deformation, in which Time Dependent Notch Sensitivity was eliminated. In every case the theory was in complete agreement with the experimental results of reference (6). The precipitate cracking characterized heretofore, when combined at identical locations with high temperature creep deformation is the nucleation mechanism for premature crack initiation.

D. EFFECT OF SOLUTION TEMPERATURE

Decreasing the solution treatment temperature without changing the solution time, keeping the aging time and temperature constant, decreases the amount of alloying elements available in the solution treated and quenched condition.

This decreases the amount of precipitates available during aging, and decreases the precipitate size at the critical distance. However, the reduction in critical size does not effect the critical distance between spherical coherent precipitates and the maximum yield point associated with this distance.

Let's compare the behavior of Alloy 718, when the aging treatment is held fixed at 48 hours and 1350°F and the solution time held constant at 10 hours (Figure 6). When the material is solution treated at 1950°F and creep rupture tested at 1000°F, (Figure 6) the distance between precipitates is 39 Burgers vectors. The specimen exhibits severe Time Dependent Notch Sensitivity, at a shear yield point of 59 ksi[6]. Reducing the solution temperature to 1800°F reduces the alloying elements available for precipitation and this reduces the critical particle size. Since the average distance between precipitates Ω is closer now than when solutioned at 1950°F to the critical distance, there will be a small increase in the particle size, resulting in an increase in the distance between precipitates from 39 to 41 Burgers Vectors, with a corresponding increase in the yield strength to 61 ksi. Here, Ω is closer to Ω critical than at the higher solution temperature and Time Dependent Notch Sensitivity has been reduced but not eliminated. If the aging process is now applied with further reduction in solution temperature a critical dislocation line length is reached, together with its corresponding maximum yield point but with a smaller critical particle size.

Finally, by reducing the solution temperature to 1700°F the critical precipitate size is further reduced, and is surpassed by the aging treatment. Here the length of dislocation line segment goes beyond the critical length of 50 to 167 Burgers vectors with a corresponding reduction in yield strength from the critical value of 68 ksi to 58 ksi. This however satisfies the specification requirements set forth in AMS 5596 for plane stress deformation. The material is now overaged, and Orowans criteria(3), set forth in equation(6) is satisfied. The dislocation line segments now pass between precipitate particles without shearing them and Notch Induced Microcrack Initiation is completely eliminated.

Hence, reducing the solution treating temperature, decreases the amount of alloying elements available in solid solution, which decreases the precipitate size at the critical distance. This does not alter the critical distance itself nor the maximum yield strength which occurs at the critical distance, when the dislocation line segment between spherical coherent precipitates, is bent to its maximum radius of curvature.

E. THE TRANSFORMATION OF GAMMA DOUBLE PRIME TO DELTA

The nucleation of the δ needles appear to be connected with the occurrence of stacking faults which are frequently observed within γ'' plates(7). This depletes the alloy of the major hardening phase in Alloy 718.

Figure (8) is an optical micrograph taken from production parts that were given the proposed heat treatment. Note that the majority of the δ phase exists as grain boundary globular δ, while most of the Widmanstätten needle structure has been eliminated. This is because this solution treatment is applied for 10 hours at 1700°F and is sufficient to dissolve the detrimental δ platelets that transformed from γ'' during high temperature service. The existing GE procedure B50T69B calls for 1 hour solution time at 1700°F which is inadequate for diffusion of the platelets into solution.

The present work presented heretofore has been one of the first applications, where dislocation theory was used to directly solve a reliability problem in production. The dynamic reactions that occurred at the atomic level, as the parameters of dispersion size Ω, Burgers vector b, particle diameter d, related the theory of the flexibility of the dislocation line segment with the effect of the flow stress, to determine if the dislocation will cut or pass between precipitates, in order to eliminate Time Dependent Notch Sensitivity, and intimately effect the serviceability of the final part.

F. CONCLUSIONS

1. Time Dependent Notch Sensitivity is a primary cause of premature high temperature crack initiation, occurring in alloy 718. This shortens the service life of engine components. Laboratory tests on the proposed heat treatment have proven to completely eliminate Time Dependent Notch Sensitivity from this alloy. This resulted in an increase of the service life, and in an improved material.

2. The in-service transformation from gamma double prime to delta platelets in alloy 718 lowers the yield point strength by depleting the precipitate and produces stress concentrations within the grains to adversely effect the ductility and fracture characteristics of the alloy. A microstructural analysis of a service part heat treated by the proposed procedure, shows a substantial reduction of the delta phase, as compared to the microstructure shortly after removal from service, while the tensile strength of the heat treated item remained within specification limits.

3. Time Dependent Notch Sensitivity is due to the way dislocations break from their precipitate barriers in reference to a average critical distance between the precipitates. When the average distance between precipitates is less than or equal to the critical distance, the dislocations will shear through the coherent particle, creating localized plastic deformation. When this is combined at identical locations with high temperature creep deformation a crack will nucleate prematurely.

4. When the average distance between precipitates is greater than the critical distance, the dislocations will pass between the precipitates without cracking them and Time Dependent Notch Sensitivity does not occur.

5. Decreasing the amount of precipitate in solid solution lowers the precipitate size at the critical distance, but does not alter the critical distance itself, or its maximum yield point. It is this critical distance that must be exceeded in overaging spherical coherent precipitates, in order that the dislocations pass between them.

6. The results of hardness testing are inadequate to accurately evaluate the heat treatment of a precipitation hardened industrial alloy. Hardness testing should be replaced by a static tensile or shear yield point test, combined with a calculation of the number of atomic distances between precipitates.

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