Strengthening mechanisms in Inconel 718 superalloy

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The yielding behaviour of underaged Inconel alloy 718, which is precipitation hardened by coherent ordered bct γ'' and ordered fcc γ' phases, has been studied. It was found that the yielding of the underaged specimens occurs by the shearing of the precipitate particles by the glide dislocations which move in pairs. During this stage the yield strength of material can be adequately accounted for mainly by the coherency hardening mechanism. MS/0798

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The main strengthening mechanism in a number of underaged nickel containing alloys, strengthened by the presence of coherent γ' precipitates, which have an ordered L12 type structure, has been attributed to the order hardening mechanism.[1–8] However, in the near peak aged condition the coherency hardening mechanism also seems to make a significant contribution.[8,9] In Inconel 718* the major precipitating phase has been reported to be γ'', i.e. in the peak aged condition the volume fraction of γ'' phase has been observed to be ~15% while that of γ' phase was ~4%.[10] Although the γ'' phase also has an ordered structure, which is DO_{22} type bct, and precipitates coherently in the fcc matrix, the strengthening in the near peak aged Inconel 718 has been reported to be principally due to the coherency strain hardening mechanism.[11] However, recently, the yield strength of an underaged Co–Ni–Cr base alloy, where only γ'' phase was present, was attributed to both the order and coherency strain hardening mechanisms.[12] Therefore, a project was initiated to further investigate the strengthening mechanisms in Inconel alloy 718. The influence of various aging heat treatments on the size and distribution of γ' and γ'' precipitates in this alloy has been published earlier.[13] This paper is concerned with the mechanical properties and precipitation strengthening mechanisms in underaged Inconel alloy 718.

EXPERIMENTAL TECHNIQUES

A 5·0 mm thick sheet of commercial grade Inconel 718, whose chemical composition is listed in Table 1, was supplied by the International Nickel Company of Canada. This sheet was cold rolled to 1·5 and 0·2 mm thick strips, with several intermediate anneals at 1473 K. The 1·5 mm thick strip was used for making 20 mm gauge length tensile specimens with a 1·5 × 6·0 mm cross-section. The 0·2 mm thick strips were used for making thin foils for electron microscopy. The tensile and 0·2 mm thick strip specimens were given a final solution heat treatment for 1 h at 1473 K and quenched in ice water. They were then aged for various lengths of time at temperatures of 973, 998, and 1023 K. All the heat treatments were carried out with specimens sealed in Vycor capsules partially filled with argon gas. On water quenching the capsules were always broken under water to attain a rapid cooling rate. The structure of aged as well as aged and deformed specimens was examined by thin film electron microscopy. The thin films were prepared by electropolishing 3·0 mm dia. discs of heat treated 0·2 mm strips in a jet electropolishing unit using a 15% perchloric acid–85% methanol bath at 223–233 K. The thin films were examined in a Philips 300 electron microscope. The experimental details of particle size measurements of γ'' and γ' precipitates have been presented elsewhere.[13] The volume fraction of the precipitate was measured by the electrochemical extraction technique the details of which have also been presented elsewhere.[13]

* Inconel is a trade mark of Huntington Alloys Inc.

RESULTS

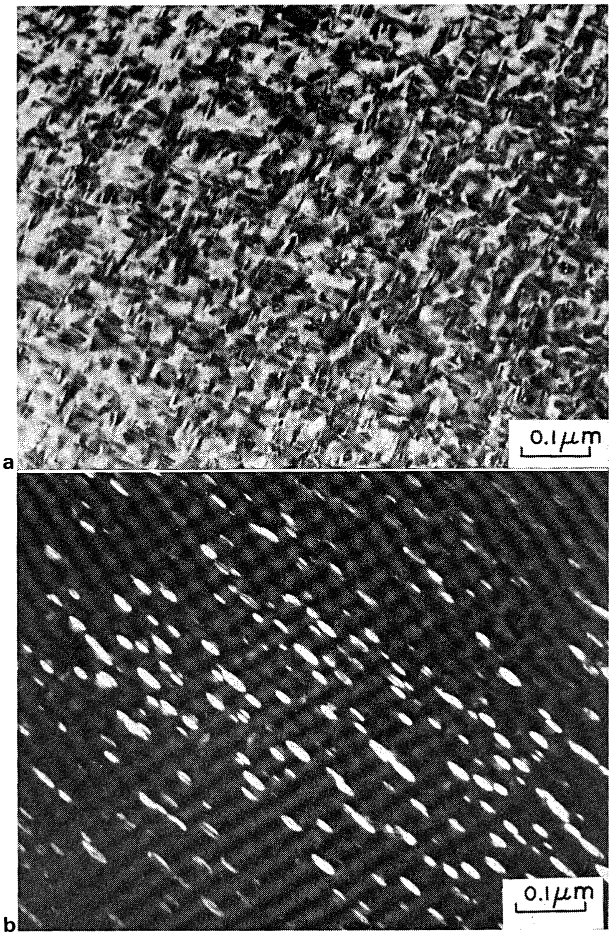
Precipitation of γ'' and γ' phases

The precipitation of γ'' and γ' phases observed by several investigators[10,11,13,14] during the early stage of aging of Inconel 718 was confirmed during the present investigation. As reported elsewhere,[10] it was found that the γ'' phase precipitates coherently on {100} planes of the fcc matrix and has a disc shape with the *c* axis being perpendicular to the discs. The γ' phase was observed to be spherical and has the same crystallographic habit as observed in many other alloy systems.[6,10,11,13,15] A typical aged structure is shown in Fig. 1*a* which is the structure of a specimen aged for 5 h at 1023 K. The orientation of the foil is {100} and its dark field taken with a $(100)(\gamma'+\gamma'')$ reflection is shown in Fig. 1*b*. In this micrograph the spherical particles are γ' and the plate shaped particles are γ''. The lattice misfits between γ' and γ'' and the matrix were found to be 0·407 and 2·86% respectively.

The volume fraction of both γ'' and γ' phases was measured by electrochemical extraction of the precipitate particles from the grip section of the used tensile specimens. By this technique only the combined volume fraction of both the phases could be determined. The variation in this volume fraction with aging time at various temperatures is shown in Fig. 2. It is observed that at all three aging temperatures the volume fraction first increases and then acquires a constant value. The aging time required to achieve the constancy of volume fraction and the value of this volume fraction increase with a decrease in temperature. The ratio of the volume fractions

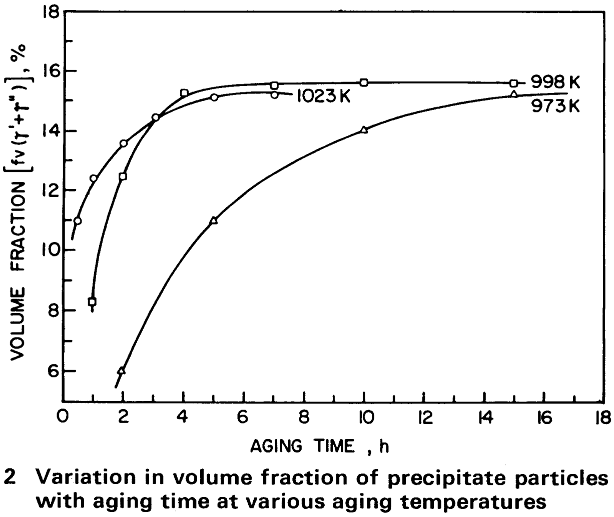
Table 1 Chemical composition of Inconel 718, wt-%

Ni	Fe	Cr	Nb	Al	Ti	Mo	C
52·37	19·24	18·24	4·94	0·52	0·97	3·07	0·03

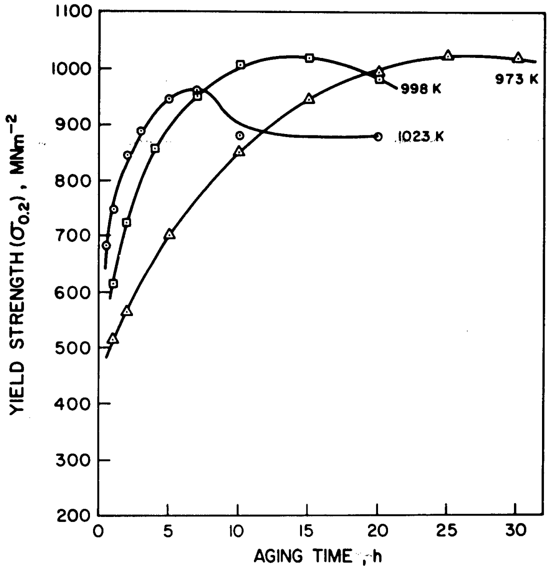


1 *a* typical structure of specimen aged for 5 h at 1023 K; *b* dark field structure of *a* taken with (100)($\gamma'+\gamma''$) reflection

of γ'' and γ' phases was measured for a number of aged specimens by dark field electron microscopy. For each specimen it was found to be between 2·5 and 3·5, which is in good agreement with the earlier reported[10] value of 3·0. Therefore, the individual volume fractions of γ'' and γ' were obtained by assuming their ratio to be 3·0. It was not possible to measure the size of extremely fine precipitate particles present during the very early stages of aging. Therefore, their sizes were estimated by extrapolating the particle size–aging time plots published earlier.[13]



2 Variation in volume fraction of precipitate particles with aging time at various aging temperatures



3 Variation in 0·2% yield strength with aging time at various aging temperatures

Tensile test results

The influence of aging time on the yield strength of specimens aged at various temperatures was established by room temperature tensile testing. The tensile specimens were tested in an Instron machine at an initial strain rate of $4{\cdot}16\times10^{-4}$ s^{-1}. The variation in 0·2% yield strength with aging time at 973, 998, and 1023 K is shown in Fig. 3. It is observed to follow a normal behaviour, i.e. as the temperature of aging is increased the yield strength of the peak aged specimen, and the time required to attain it, are decreased.

To determine the temperature dependence of the yield strength specimens aged for 5·5 h at 998 K, along with solution treated specimens, were tested at 77, 183, and 293 K. The values of elastic moduli at these temperatures were determined by monitoring the elongation with strain gauges attached to the specimens. The results of these tests are given in Table 2. It is observed that the value of the precipitation hardening component of yield strength $\Delta\sigma$ decreases as the testing temperature is decreased. If these values are normalized by dividing them by the appropriate values of elastic modulus E at the test temperature, the influence of temperature becomes even more pronounced. That is, as the temperature of testing is decreased from 293 to 77 K the value of $\Delta\sigma/E$ is seen to decrease by about 40%. To investigate the influence of test temperature further, two sets of specimens aged for 2 h at 1023 K and 10 h at 973 K were also tested at 293 and 77 K. The results obtained are listed in Table 3. Once again it is seen that a significant reduction in the precipitation hardening component of yield strength $\Delta\sigma/E$ occurred as the test temperature decreased from 293 to 77 K.

Deformation structures

In order to study the deformation behaviour of this alloy the dislocation–precipitate interactions were studied by thin film electron microscopy. The tensile specimens of 0·2 mm thick strips were aged and then deformed 3–5% in an Instron machine similar to a regular tension test. Thin foils from the gauge section of the deformed specimen were then examined in an electron microscope. It was observed that the presence of a very high level of coherency strain around γ'' and γ' precipitate particles tended to mask the dislocation contrast completely. However, this effect was less pronounced in specimens aged for a very short aging time. A large range of tilt angles was also used to suppress

Table 2 Effect of test temperature on yield strengths of specimens solution treated and aged for 5·5 h at 998 K

Temperature, K	0·2% yield strength of aged specimens, MN m^{-2}	Young's modulus E, GN m^{-2}	0·2% yield strength of solution treated specimen, MN m^{-2}	$\Delta\sigma$, MN m^{-2}	$\Delta\sigma/E \times 10^{-3}$
77	996	23·17	489	507	21·79
183	925	22·80	364	561	24·60
293	849	20·04	241	608	30·34

Table 3 Effect of test temperature on yield strength of aged specimens

	Test temperature, K						
	293			77			
Treatment	$\sigma_{0\cdot2}$	$\Delta\sigma$	$\Delta\sigma/E$	$\sigma_{0\cdot2}$	$\Delta\sigma$	$\Delta\sigma/E$	$\frac{(\Delta\sigma/E)_{293}}{(\Delta\sigma/E)_{77}}$
2·0 h at 1023 K	813	572	28·5	965	476	20·63	1·38
5·5 h at 998 K	849	608	30·3	996	507	21·88	1·39
10 h at 973 K	751	510	25·4	900	411	17·73	1·43

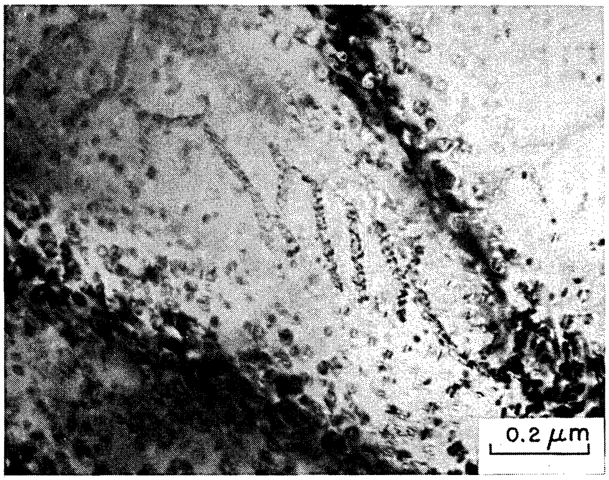
the coherency strain field contrast and enhance the dislocation contrast. Figure 4 shows the structure of a specimen aged for 10 min at 1023 K and deformed 3%. The dislocations are seen to occur in pairs. Another example of the pairwise motion of dislocations on deformation is shown in Fig. 5, which is the structure of a specimen aged for 30 min at 1023 K. The diameters of γ'' and γ' precipitate particles at this stage of aging are estimated to be 37 and 100 Å respectively. When an ordered particle is sheared by a glide dislocation an antiphase boundary is created within the particle. This causes a second dislocation to be pulled forward removing the disorder within the particle. Therefore, whenever ordered particles are sheared dislocations are seen to occur in pairs. Therefore, the presence of dislocation pairs in the aged and deformed specimens suggests that the deformation of underaged specimens of Inconel 718 occurs by the shearing of the ordered precipitate particle by glide dislocations. This is similar to the precipitate–dislocation interaction observed in other γ' and γ'' precipitation hardened alloys.[1,4–9,11,12]

DISCUSSION

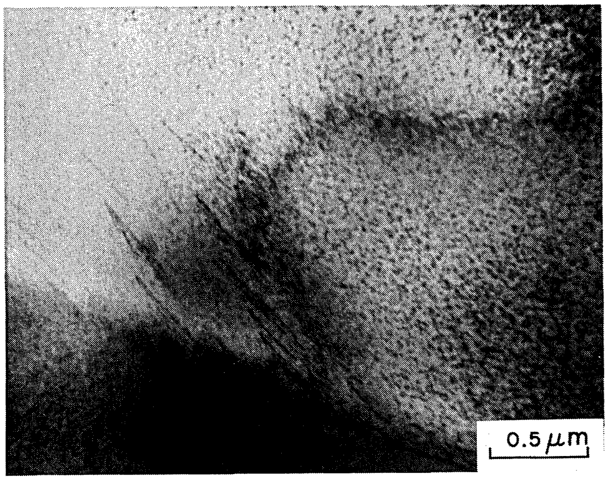
The above results show that during the early stages of aging of Inconel 718 the prominent precipitating phase is γ'', but a certain amount of γ' phase also forms. During this stage the deformation of aged material occurs by the pairwise motion of dislocations, which suggests that the ordered precipitate particles are being sheared by the glide dislocations. In general, when precipitate particles are sheared by dislocations during deformation the strengthening of the alloy can arise due to (a) coherency strain hardening, (b) order hardening, (c) modulus hardening, (d) surface hardening, and (e) stacking fault hardening mechanisms. It has been observed that in nickel alloys precipitation strengthened by γ' the strengthening is mainly due to order and coherency hardening mechanisms and the contribution by other mechanisms is almost negligible.[4,5] Since, like γ' precipitates γ'' phase also has an ordered structure and precipitates coherently, the strengthening of the underaged Inconel 718 should also be due to the order and/or coherency strengthening mechanisms.

In order to determine which of these two mechanisms controls the yielding behaviour and strength of the underaged Inconel 718 their contribution to $\Delta\tau$, the critical resolved shear stress (CRSS), is first calculated. The theoretically calculated values for each of the specimens are then compared with the observed values of CRSS. Since a significant amount of γ' phase is also present, the overall strength should depend upon the contributions to CRSS by both γ' and γ'' phases. It has been suggested that when a combination of strong and weak obstacles is present, their combined influence on CRSS is more accurately given by the mean square law,[16] i.e.

$$\Delta\tau_{\text{total}} = (\Delta\tau_{\gamma'}^2 + \Delta\tau_{\gamma''}^2)^{1/2}$$



4 Structure of specimen aged for 10 min at 1023 K and 3% deformed



5 Structure of specimen aged for 30 min at 1023 K and 3% deformed

Order strengthening

The increment in CRSS due to order hardening when disc shaped ordered bct particles γ'' are sheared by the pairwise motion of glide dislocation $\Delta\tau_{0\gamma''}$ has been calculated to be given by[11]

$$\Delta\tau_{0\gamma''} = \frac{\gamma_{APB''}}{2b}\left\{\left[\frac{4\gamma_{APB''}f_{v''}}{\pi T}\left(\frac{\sqrt{6}Rh}{3}\right)^{1/2}\right] - \beta f_{v''}\right\}$$

where $\gamma_{APB''}$ is the antiphase boundary energy, $f_{v''}$ is the volume fraction, R is the radius, and h is the half thickness of γ'' particles, while T and b are the line tension and Burgers vector of the glide dislocation, respectively. The value of T for an edge dislocation is assumed to be equal to $\mu b^2/2$, where μ is the shear modulus. The value of constant β depends upon the distribution of γ'' precipitate and is equal to 1/3 when they are present along all the three invariants,[11] as observed in the present investigation.

When the sheared particles are spherical shape ordered γ', the contribution to CRSS by the order hardening mechanism $\Delta\tau_{0\gamma'}$ is given by[17]

$$\Delta\tau_{0\gamma'} = \frac{\gamma_{APB'}}{2b}\left[\left(\frac{4\gamma_{APB'}f_{v'}R_s}{\pi T}\right)^{1/2} - f_{v'}\right]$$

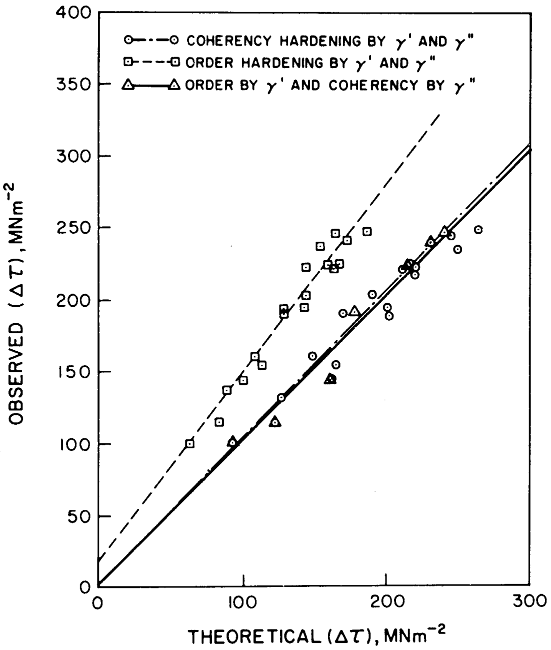
where $\gamma_{APB'}$ is the antiphase boundary energy, $f_{v'}$ is the volume fraction, and R_s is the effective radius of γ' precipitate particles.

Oblak *et al.*[11] have calculated the antiphase boundary energies of γ'' and γ' phases in Inconel 718 to be 296 and 12 mJ m^{-2}. The value for γ' phase is extremely small compared with the generally quoted[18] values of between 150 and 190 mJ m^{-2}. Furthermore, this value of $\gamma_{APB'}$ yields a negative value of $\Delta\tau_{0\gamma'}$ when used to calculate the order strengthening by γ' phase in underaged specimens, since the value of R_s is also small. In a Ni–Co–Cr–Nb–Al alloy,[19] where like Inconel 718 the γ' phase is also $Ni_3(NbAl)$ type, the value of $\gamma_{APB'}$ was found to be 175 mJ m^{-2}. Therefore, it seems more appropriate to use a value of 175 mJ m^{-2} for the antiphase boundary energy for γ'. Therefore, the antiphase boundary energy of 296 mJ m^{-2} for γ'' and 175 mJ m^{-2} for γ' phase was used in the present studies. The theoretical values of $\Delta\tau_{0\gamma''}$ and $\Delta\tau_{0\gamma'}$ for variously aged specimens were calculated by using these values of antiphase boundary energy, the observed particle sizes, and their volume fraction. The actual contribution of precipitation strengthening to the yield strength was determined by the expression $(\sigma_y - \sigma_0)$, where σ_y and σ_0 are the 0·2% yield strengths of the aged and solution treated specimens, respectively. The increase in CRSS, $\Delta\tau$ observed, was then obtained by dividing $(\sigma_y - \sigma_0)$ by 3·1, which is the Taylor factor for the fcc materials. The observed values of $\Delta\tau$ were then plotted against the theoretical values due to order strengthening, both by the γ' and γ'' precipitate particles, as shown in Fig. 6. If the order strengthening mechanism is responsible for the observed increase in strength this should be a straight line with a slope of 1·0 and should pass through the origin. The slope of the best fitting line was found to be 1·290±0·075 and the value of the y axis intercept was 19.15±10.4 MN m^{-2}. Therefore, it is seen that the order strengthening underestimates the value of CRSS by about 30%.

Coherency strengthening

The contribution of coherency strain hardening by the γ' precipitate particles in an fcc matrix $\Delta\tau_{c\gamma'}$ can be calculated by the following expression due to Gerold and Haberkorn[20]

$$\Delta\tau_{c\gamma'} = \alpha\mu|\varepsilon_{\gamma'}|^{3/2}\left(\frac{R_s}{b}\right)^{1/2} f_{v'}^{1/2}$$



6 Plots of observed *v.* theoretical values of $\Delta\tau$ during early stages of aging

where $\varepsilon_{\gamma'}$ is the γ' matrix misfit parameter and α is a constant which has values of 3 and 1 for edge and screw dislocations, respectively. The other terms are the same as defined earlier. The coherency hardening by the disc shaped bct precipitate γ'' in an fcc matrix, when the glide dislocations are of an edge, can be obtained by[11,21]

$$\Delta\tau_{c\gamma''} = 1{\cdot}7\mu|\varepsilon_{\gamma''}|^{3/2}\left(h^2\frac{f_{v''}(1-\beta)}{2bR}\right)^{1/2}$$

where $\varepsilon_{\gamma''}$ is the γ'' matrix mismatch parameter.

The theoretical values of $\Delta\tau_{c\gamma'}$ and $\Delta\tau_{c\gamma''}$ for variously aged specimens were determined by using the observed values of volume fraction, particle size, and misfit parameter. The total theoretical values of CRSS due to the coherency hardening mechanism was also obtained by the mean square law method. The theoretical values were then once again plotted against the observed values of $\Delta\tau$, as shown in Fig. 6. The slope of the best fit straight line was found to be 1·005±0·05 and the value of the y axis intercept was 1·79±10·5 MN m^{-2}. It is seen that the values of the CRSS predicted by the coherency strain hardening mechanism are in excellent agreement with the observed values. Therefore, it would seem that the yielding behaviour and yield strength of the underaged Inconel 718 is controlled by the coherency strain hardening mechanism.

It should be noted that in a number of underaged γ' precipitation hardened alloys the order hardening is found to be the strengthening mechanism.[4–7] The above analysis, however, has considered the strengthening contribution by the γ' precipitate particles, although relatively small, by the coherency strengthening mechanism. It is also possible that the contribution by the γ' phase is due to the order hardening mechanism instead. To evaluate this possibility the total theoretical value of CRSS, by the combined influence of coherency hardening by γ'' and order hardening by γ', was also calculated. Once again the mean square law method was used and the calculated values of $\Delta\tau$ were plotted against the observed values. This plot, also shown in Fig. 6, has a slope 1·021±0·059 and the value of the y intercept is 2·95±11·24 MN m^{-2}. This suggests that the theoretical values of CRSS predicted by a combination of order hardening by γ' and coherency hardening by γ'' are also in excellent agreement with the observed values. Therefore, it is not possible to establish whether the contribution to CRSS by γ', although relatively small, is by the order hardening or by the coherency hardening mechanism. However, it does seem that the major strengthening is provided by the coherency hardening mechanism by γ''.

Indirect evidence to support the above conclusion may also be found in the temperature dependence of $\Delta\tau$. In the peak hardened Cu–Co system[22] where the strengthening is entirely due to the coherency hardening mechanisms, as the testing temperature is decreased from 303 to 77 K the value of flow stress is also seen to decrease in proportion to the reduction in the value of ε. The variation in the value of ε with temperature in Inconel 718 is unfortunately not known. However, in the Ni–Al system,[5] which is precipitation hardened by Ni_3Al γ' precipitate, the value of ε decreases from 0·355% at room temperature to 0·255% at 77 K. Therefore, it is likely that in the present nickel base alloy the value of ε will also decrease as the temperature is reduced from 293 to 77 K since both γ' and γ'' are of the $Ni_3(AlTi)$ type of compounds. If the coherency strain hardening mechanism were responsible for the precipitation hardening, this would explain the observed reduction in the value of $\Delta\tau/E$ as the temperature is reduced from 293 to 77 K.

CONCLUSIONS

1. During the early stages of aging the strength of Inconel 718 is attributed to the coherent precipitation of ordered bct γ'' and ordered fcc γ' phases. The ratio of γ'' to γ' phase was 3·0 to 1·0.
2. Yielding of the underaged material occurs by the shearing of the precipitate particles by the glide dislocations which move in pairs.
3. The strength of the underaged alloy can be adequately accounted for by coherency strain hardening by γ'' and γ' particles; the strengthening by γ' particles could be by order hardening instead.

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