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A Study of Back Stress During Creep Deformation of a Superalloy Inconel 718

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ABSTRACT

The values of back stress during creep deformation of Inconel 718 superalloy have been determined by the consecutive stress reduction method in the stress range 620-865 MN m^{-2} and in the temperature range 853-943 K. The results show that the values of back stress are as high as (0.58-0.97)σ_a where σ_a is the applied stress. In the low stress region ($\sigma_a \leqslant 720$ MN m^{-2}) where diffusional creep controls the deformation, a threshold stress $\sigma_{0(th)}$ is observed to exist. In the high stress region ($\sigma_a \geqslant 746$ MN m^{-2}) where power law creep is observed to be operative, the back stress σ_0 is proportional to the applied stress σ_a. The rate of variation in σ_0 with σ_a, i.e. the value of dσ_0/dσ_a, *was found to be about 0.25 at temperatures in the range 873-898 K.*

The value of threshold stress $\sigma_{0(th)}$ was dependent on the particle size $d_{\gamma''}$. At 873 K and stresses below 720 MN m^{-2}, $\sigma_{0(th)}$ increases with an increase in particle size until a maximum value is reached. $\sigma_{0(th)}$ then decreases with further increase in particle size. Comparison of the experimentally measured threshold stress $\sigma_{0(th)}$ with the theoretically calculated value of σ_0 suggests that, in the small-particle-size region, Harris' model involving the formation and punching of prismatic dislocation loops controls the diffusional creep process, while in the large-particle-size region, Ashby's model of grain boundary edge dislocation climb is rate controlling.

1. INTRODUCTION

During steady state creep deformation of two-phase alloys the action of the applied stress is opposed by a back stress resulting from the presence of second-phase particles and a defect structure within the material. Therefore, creep deformation is the result of an effective stress which is given by $\sigma_a - \sigma_0$ where σ_a is the applied stress and σ_0 is the back stress. As a result, in the steady state creep rate expression, σ_a is replaced by $\sigma_a - \sigma_0$, *i.e.* the steady state creep rate $\dot{\epsilon}_s$ is given by

$$\dot{\epsilon}_s = A^*(\sigma_a - \sigma_0)^{n_e} \exp\left(-\frac{Q_e}{RT}\right) \qquad (1)$$

where n_e is the effective stress exponent whose value is 1 for diffusional creep and between 3 and 6 for dislocation power law creep, Q_e is the effective activation energy for the creep process (often equal to the activation energy for self-diffusion), R is the gas constant and T is the absolute temperature. Therefore, in order to study the creep deformation behaviour and creep mechanism in two-phase alloys, it is desirable to determine the values of back stress.

The commercial superalloy Inconel 718, which is precipitation hardened by the presence of 13 vol.% of coherent ordered body-centred tetragonal γ'' and 4 vol.% of coherent f.c.c. γ' precipitates, is widely used in high temperature applications. However, the influence of the size of γ'' and γ' precipitate particles on creep deformation and the creep mechanism has not been reported. Therefore, a project was initiated to study this aspect of creep deformation of Inconel 718. In this paper the influence of particle size on the values of back stress under various creep-testing conditions is presented.

2. EXPERIMENTAL TECHNIQUES

The material used in this investigation was commercial alloy Inconel 718 provided by the International Nickel Company of Canada.

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TABLE 1

Chemical composition of Inconel 718

Element	C	Fe	Ni	Cr	Al	Ti	Mo	Nb + Ta	Mn	S	Si	Cu
Amount (wt.%)	0.03	19.24	52.37	18.24	0.52	0.97	3.07	4.94	0.007	0.007	0.30	0.04

The chemical composition of the alloy is shown in Table 1. Flat tensile creep specimens with a cross section of 1.3 mm × 5.3 mm and a gauge length of 25.4 mm were machined from strips 1.3 mm thick cold rolled from sheets 3.4 mm thick of as-received material. These specimens were solution treated at 1323 K for 1 h in pure argon and quenched in cold water. They were then given an aging treatment at 998 K for various lengths of time and then quenched in cold water to obtain various particle sizes of γ'' and γ' phases.

The tensile creep tests were carried out using two T48 Avery–Denison constant-stress creep machines with the specimens enclosed in argon-filled chambers. The creep strain was monitored during tests by measuring the displacement of an extensometer attached to the specimen grips using a linear variable-differential transducer linked to a strip chart recorder. The testing temperature was controlled to within ± 2 K with a three-zone furnace and monitored with two thermocouples attached to the upper and lower gauge length portions of the specimen.

The back stress or threshold stress was determined by the consecutive stress reduction method [1–3]. When the creep rate for a given applied stress remained constant for 5–10 h, it was assumed that steady state creep has been achieved and the applied stress was reduced by a small amount $\Delta\sigma_1 (\approx 0.05\sigma_a)$. This stress reduction resulted in an elastic contraction of the specimen, which was followed by an incubation period of zero creep rate. Creep then restarted with a slower creep rate at the reduced stress level. When a new steady state creep rate was reached, a second small stress reduction $\Delta\sigma_2$ was made, and so on. The duration of successive incubation periods Δt_1, Δt_2, Δt_3, ... were recorded for each consecutive small stress reduction. To determine σ_0, the data were plotted as cumulative incubation time $\Sigma\Delta t$ against the remaining stress on a linear scale. The back stress or threshold stress was taken as the asymptotic value of the remaining stress when the cumulative incubation time appeared to be infinite.

The experimental details for measuring the particle size of γ'' and γ' precipitates have been reported elsewhere [4]. Since the coarsening rates of γ'' and γ' precipitates follow the Lifshitz–Wagner theory of diffusion-controlled growth, *i.e.* $\bar{d}^3 - \bar{d}_0^3 = Kt$ [4], the particle sizes in the underaged condition which are too small to be measured microscopically were obtained by extrapolation.

The microstructures of aged and crept specimens were examined by optical and thin foil electron microscopy. The etchant used for optical microscopy was a solution of 1 part of 10% Cr_2O_3 and 3 parts of HCl. To prepare thin foils for electron microscopy, gauge length parts 1.3 mm thick of the crept specimens were hand ground to strips about 0.2 mm thick and discs 3.0 mm in diameter were then punched from them. These 3.0 mm discs were electropolished in a jet electropolishing unit using a 15% perchloric acid–85% methanol bath at 223–233 K. The thin foils were examined in a Philips 300 electron microscope.

3. RESULTS

The values of back stress for underaged, peak-aged and overaged specimens were determined at applied stresses in the range 620–865 MN m^{-2} and at temperatures in the range 853–943 K. The strain *versus* time curves showed the presence of incubation periods after small stress reductions, the length of which depended on the testing conditions, *i.e.* the remaining stress, the testing temperature and the aging time of the specimens. A typical example is illustrated in Fig. 1, which is the creep curve of a specimen aged at 998 K for 15 h and tested at 873 K with an initial applied stress of 790 MN m^{-2}.

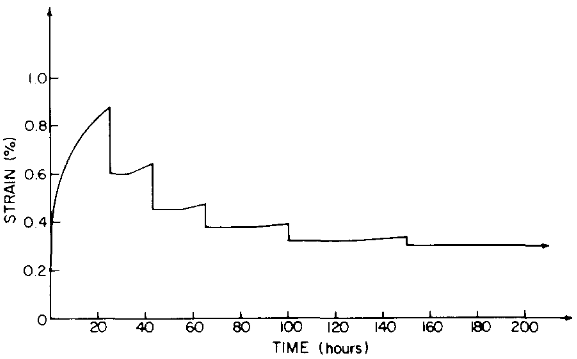

Fig. 1. Strain *vs.* time curve during a consecutive stress reduction test. The specimen was aged at 998 K for 15 h and tested at 873 K and a stress of 790–622 MN m^{-2}.

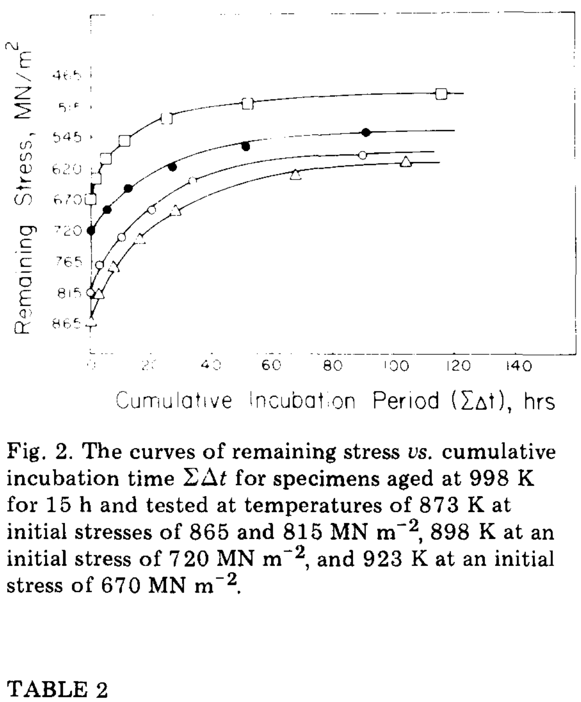

Fig. 2. The curves of remaining stress *vs.* cumulative incubation time $\Sigma\Delta t$ for specimens aged at 998 K for 15 h and tested at temperatures of 873 K at initial stresses of 865 and 815 MN m^{-2}, 898 K at an initial stress of 720 MN m^{-2}, and 923 K at an initial stress of 670 MN m^{-2}.

The amount of each stress reduction was 5% of the initial applied stress. To determine the back stress, the curves of the remaining stress *versus* cumulative incubation period $\Sigma\Delta t$ were plotted. A few typical examples of these curves are shown in Fig. 2. The back stress is taken as the asymptotic value of remaining stress when the cumulative incubation time appears to be infinite.

The experimentally measured values of σ_0 for various specimens are given in Table 2. It is seen that the values of back stress of the present alloy are in the range $(0.58\text{–}0.97)\sigma_a$. The results in Table 2 show that, when specimens aged for 15 h at 998 K were tested in the temperature range 853–923 K, the value of σ_0 remained constant, *i.e.* there exists a threshold stress $\sigma_{0(\text{th})}$ as long as the applied stresses are below a critical value σ_c. These critical values were found to depend on test temperatures. For example, as seen in Table 2, the value of σ_c at 853 K and 873 K was 746 MN m^{-2}, while at 898 K and 923 K the values of σ_c were 720 MN m^{-2} and 695 MN m^{-2} respectively. The region where the back stress is constant is considered to be the diffusional creep region [5]. However, when the applied stress exceeded these critical stresses, the back stress increased with applied stress. In this region, creep deformation was observed to follow the power law creep mechanism [5]. The rate of increase in σ_0 with applied stress was found to be about 0.25 at temperatures of 873 and 898 K. The results listed in Table 2 also show that both the threshold stress $\sigma_{0(\text{th})}$ and the back stress σ_0 decreased with

TABLE 2

Experimentally determined back stress of Inconel 718 alloy

Aging time at 998 K (h)	$d_{\gamma''}$ (nm)	*Testing temperature* (K)	σ_a (MN m^{-2})	σ_0 (MN m^{-2})	σ_0/σ_a
15	26.9	853	643, 670, 746	617	0.83–0.95
15	26.9	873	620, 670, 720	576	0.80–0.93
15	26.9	873	765	588	0.77
15	26.9	873	815	604	0.74
15	26.9	873	865	615	0.71
15	26.9	898	620, 670, 720	540	0.75–0.87
15	26.9	898	765	571	0.74
15	26.9	898	815	583	0.71
15	26.9	898	840	589	0.70
15	26.9	923	620, 670, 695	510	0.73–0.82
15	26.9	943	620, 670, 695	437, 453, 473	0.68–0.70

increasing testing temperature. In the lower stress range, *i.e.* the diffusional creep region, the threshold stress $\sigma_{0(th)}$ has values of 617 MN m^{-2}, 576 MN m^{-2}, 540 MN m^{-2} and 510 MN m^{-2} for testing temperatures of 853 K, 873 K, 898 K and 923 K respectively. These are plotted in Fig. 3. In the high stress region, *i.e.* dislocation power law creep region, only two data points were obtained. However, they seem to follow a similar trend as well. For example, when σ_a = 815 MN m^{-2}, the value of σ_0 decreased from 604 to 583 MN m^{-2} as the testing temperature increased from 873 to 898 K.

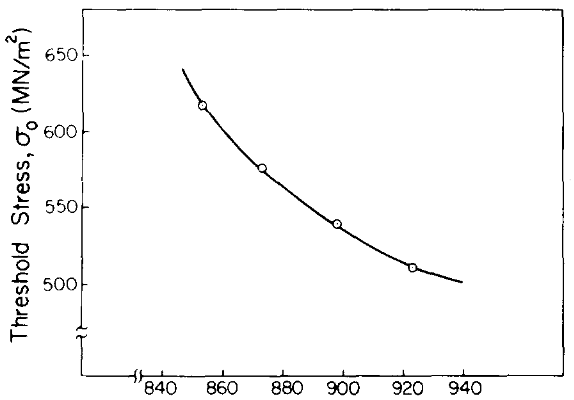

Fig. 3. The variation in threshold stress on testing temperature for specimens aged at 998 K for 15 h and tested in the stress range 620–720 MN m^{-2}.

The dependence of threshold stress $\sigma_{0(th)}$ during diffusional creep on particle size for specimens tested at 873 K and at stresses of 620–720 MN m^{-2} is also given in Table 3 and plotted in Fig. 4. This figure also shows the

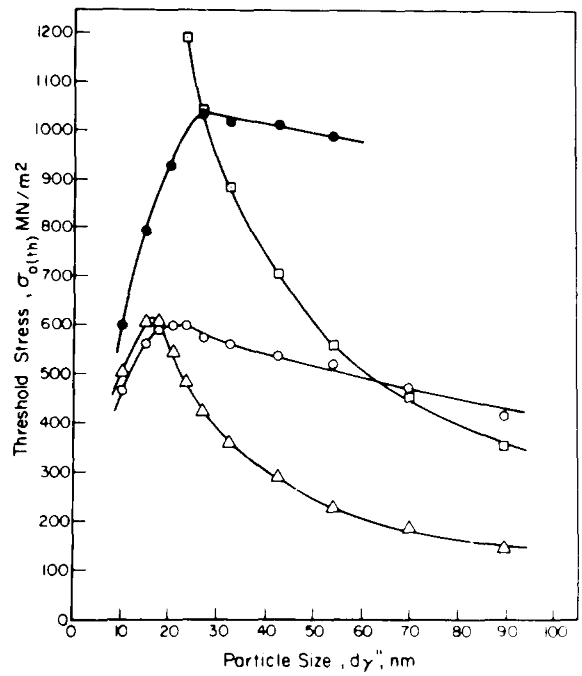

Fig. 4. The dependences of threshold stress on particle size: ⊙, measured $\sigma_{0(th)}$; △, predicted σ_{0H}; ⊡; predicted σ_{0A}; ●, $\sigma_{0.2}$ at room temperature. Specimens were aged at 998 K for 15 h and tested at 873 K and a stress of 670 MN m^{-2}.

TABLE 3

The experimental measured back stress values $\sigma_{0(th)}$, calculated value σ_{0A} from Ashby's model and calculated value σ_{0H} from Harris's model for specimens tested at 873 K and in a stress range 620–720 MN m^{-2}

Aging time at 998 K (h)	*Particle size*		*Measured* σ_0 (MN m^{-2})	*Calculated* σ_{0A} [a]	*Calculated* σ_{0H} [a]
	$d_{\gamma'}$ (nm)	$d_{\gamma''}$ (nm)			
1	6.5	9.9	467	1748	505
3	9.2	14.8	564	1626	608
5	11.0	17.2	590	1480	603
7	12.3	20.3	600	1323	540
10	13.7	23.2	600	1189	485
15	15.7	26.9	576	1037	422
25	18.4	32.6	564	885	361
50	23.0	42.4	538	708	289
100	29.2	53.8	522	558	227
200	36.0	70.2	472	452	185
400	46.0	90.8	416	353	145

[a] $\alpha = 0.6$; $b = 2.54 \times 10^{-10}$ m; G = 65.6 GN m^{-2}; $V_{t>5h} = 0.166$, $V_{t=1h} = 0.082$ and $V_{t=3h} = 0.14$ where t is the aging time at 998 K.

dependence of 0.2% yield stress at room temperature on the γ'' particle size. It is seen that, during the early stages of aging when the γ'' particle size is small, the values of $\sigma_{0(\mathrm{th})}$ and $\sigma_{0.2}$ increase with increasing particle size. However, both of them start to decrease after attaining a maximum value. The maximum in $\sigma_{0(\mathrm{th})}$ occurs at a γ'' diameter of 23 nm whereas the maximum in $\sigma_{0.2}$ is observed at 27 nm, *i.e.* the maximum in both $\sigma_{0(\mathrm{th})}$ and $\sigma_{0.2}$ seems to occur in specimens with nearly the same γ'' particle sizes. Therefore, it can be concluded that during diffusional creep of underaged material the value of threshold stress increases with increasing particle size and it decreases with increasing particle size in the overaged condition.

The dislocation structures of crept specimens consisted mainly of dislocation loops and short segments. A typical example is shown in Fig. 5 which is the micrograph of a specimen aged at 998 K for 25 h and crept at 873 K and 720 MN m^{-2} for 222 h.

4. DISCUSSION

4.1. Suitability of the consecutive stress reduction method for the determination of σ_0 during diffusional creep

According to Wilshire and coworkers [1-3] and Gibeling and Nix [6], the consecutive stress reduction method can usually be used to measure the back stress during dislocation power law creep where an incubation time

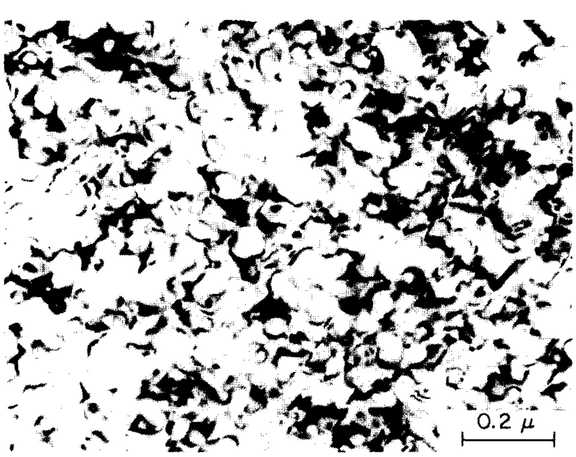

Fig. 5. Dislocation structure of a specimen aged for 25 h at 998 K and crept at 873 K and a stress of 720 MN m^{-2} for 22 h.

after a small stress reduction is observed to occur. However, in the present study an incubation time was observed in both the power law and the diffusional creep regions. Therefore, it was decided to use the consecutive stress reduction method to determine the value of back stress in both the power law and the diffusional creep regions. The reason for the appearance of an incubation time during diffusional creep may be as follows. According to Harris [7] and Burton [8], for diffusional creep to continue, the stress concentration at the precipitate-matrix interface created by the entrapment of diffusing vacancies can only be relieved by the formation of prismatic dislocation loops. However, Ansell and Weertman [9] have suggested that diffusional creep in two-phase alloys can also occur by the process of dislocation climb over the precipitate particles. Therefore, diffusional creep in two-phase alloys may involve not only vacancy diffusion in the matrix and dislocation motion in the grain boundary region but also dislocation creation and motion within grains. Similar to the process of dislocation slip during dislocation power law creep, dislocation climb during diffusional creep also requires dislocation networks or line segments to grow to a sufficient length so that they can be activated. These dislocation segments which become active are known as Bardeen-Herring sources [10]. Since the transmission electron microscopy observation of crept specimens showed the presence of dislocation networks within grains, as shown in Fig. 5, the observed incubation time could be due to the activation of these Bardeen-Herring sources. Therefore, it may be concluded that the consecutive stress reduction method can be used to determine the back stress during power law creep as well as diffusional creep in two-phase alloys.

4.2. Effect of applied stress on back stress during dislocation power law creep

The results given in Table 2 show that, at certain testing temperatures when σ_a is below a critical level, the value of σ_0 is independent of σ_a, *i.e.* there exists a threshold stress $\sigma_{0(\mathrm{th})}$. This region was observed to be the diffusional creep region [5]. However, at higher stress levels (*e.g.* for specimens aged for 15 h at 998 K, crept at 873 K and stresses higher than 765 MN m^{-2}), σ_0 increased with increasing σ_a. Theoretically, there are two approaches to predict the back stress in two-phase alloys. The threshold stress was first considered by Lund and Nix [11] and Purushothman and Tien [12] who suggested that the back stress in two-phase alloys is of the order of the Orowan stress. Nix and coworkers [6, 13] suggested a different approach for calculating the back stress. They suggested that the back stress in two-phase alloys consists of two parts, *i.e.* the back stress resulting from the interaction of moving dislocations with other dislocations, subgrain boundaries and fine twins, and the back stress arising from the presence of second-phase particles. They assumed localized dislocation climb over the particles in estimating the back stress. The former part depends on the deformation history and the applied stress σ_a. Therefore, the fact that σ_0 increases with σ_a in the high stress region in the present alloy implies that the presence of dislocation, subgrain and twin boundaries also contributes to the back stress. Since dislocation networks and deformation twins have been observed in crept specimens as shown in Fig. 6, the dependence of σ_0 on σ_a may be attributed to the interaction of a moving dislocation with other dislocations and twin boundaries as suggested by Nix and coworkers [6, 13].

4.3. *Effect of particle size on threshold stress during diffusional creep*

The results in Table 3 and Fig. 4 show that the particle size affects the back stress. In the present study the effect of particle size on back stress has been studied only in the low stress region ($\sigma_a < 720$ MN m^{-2}) where the diffusional creep process predominates. The results show that the threshold stress $\sigma_{0(\mathrm{th})}$ increases with increasing particle size in the underaged region and then decreases with further increase in particle size. There are several theoretical models to relate the threshold stress during diffusional creep to the size and volume fraction of precipitate. Ashby [14] has proposed that the diffusional creep rate is controlled by the climb of grain boundary edge dislocations whose motion is impeded by pinning points, and the value of σ_0 is given by

$$\sigma_{0\mathrm{A}} = \frac{2\Gamma}{b\lambda} = \frac{2\Gamma}{br} V^{1/2} \tag{2}$$

where Γ is the dislocation line tension, given by $\alpha G b^2$, where $\alpha = 0.5$–1, G is the shear modulus, λ is the distance between pinning points, b is the Burgers vector, r is the radius and V is the volume fraction of second-phase particles. The threshold stress $\sigma_{0(\mathrm{th})}$ in Harris' [7] model corresponds to the stress required to nucleate interstitial prismatic dislocation loops at particle–matrix interfaces and can be expressed as

$$\sigma_{0\mathrm{H}} = \frac{2\Gamma}{br} V \tag{3}$$

To determine which of these two models is applicable to the behaviour of Inconel 718, the experimentally determined values of $\sigma_{0(\mathrm{th})}$ are compared with the theoretical values in Table 3. The two values are also plotted in Fig. 4. It is seen that in the underaged condition with γ'' particle size less than 17 nm, *i.e.* when specimens were aged at 998 K for less than 7 h, the measured threshold stresses $\sigma_{0(\mathrm{th})}$ are comparable with the values predicted by the model of Harris, *i.e.* with $\sigma_{0\mathrm{H}}$. This suggests that the prismatic dislocation loop formation mechanism may determine the value of $\sigma_{0(\mathrm{th})}$ in this particle size region. When γ'' particle sizes are greater than 42 nm, *i.e.* the aging time is longer than 50 h at 998 K, the measured values of $\sigma_{0(\mathrm{th})}$ are comparable with the values predicted by Ashby's model, *i.e.* with $\sigma_{0\mathrm{A}}$. Therefore, in the large-particle-size region, grain boundary edge dislocation climb seems to determine the value of thresh-

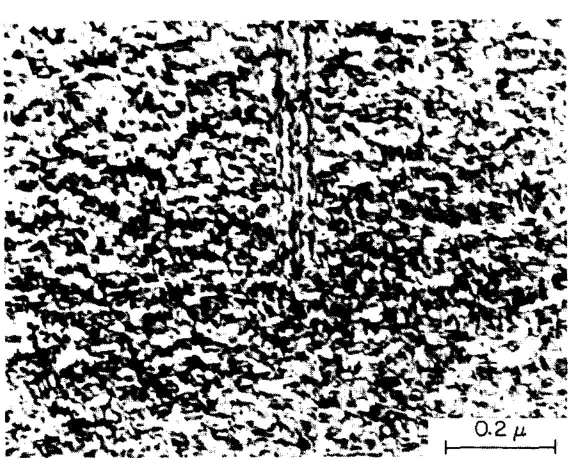

Fig. 6. Dislocation structure and fine twins of specimen aged for 7 h at 998 K and crept at 873 K and a stress of 815 MN m^{-2} for 21 h.

old stress. However, when the γ'' particle size is in the range 20-42 nm, the measured threshold stresses are comparable with neither the value of σ_{0H} nor the value of σ_{0A}. It may be considered that, in the peak-aged condition, both prismatic dislocation loop formation and grain boundary edge dislocation climb control the creep process.

4.4. *Effect of testing temperature on back stress*

The decrease in σ_0 with increasing testing temperature, as shown in Fig. 3 and Table 2, can be attributed to the dependence of shear modulus on temperature since the threshold stress predicted from all these theoretical models is dependent on the shear modulus (Γ in eqn. (2) and eqn. (3) is a function of the shear modulus). Furthermore, from an energy point of view, when the temperature increases, more thermal energy is available to assist the applied stress to activate Frank-Read and Bardeen-Herring sources or to nucleate prismatic dislocations. Therefore, σ_0 is expected to decrease with increasing testing temperature.

5. CONCLUSIONS

(1) The threshold stress $\sigma_{0(\mathrm{th})}$ during diffusional creep and the back stress σ_0 during dislocation power law creep for Inconel 718 have been found to be as high as $(0.6\text{-}0.9)\sigma_a$ in the stress range 620-865 MN m^{-2} and temperature range 853-943 K.

(2) The values of applied stress, testing temperature and γ'' and γ' precipitate particle sizes affect the magnitude of the back stress for creep of Inconel 718 superalloy.

(a) In the diffusional creep region, there exists a threshold stress $\sigma_{0(\mathrm{th})}$ which is independent of the applied stress and, in the dislocation power law creep region, the back stress is proportional to the applied stress.

(b) The back stress decreases with increase in temperature, which is attributed to the dependence of the shear modulus on temperature.

(c) The threshold stress increases with increasing particle size in the underaged condition and decreases with increasing particle size in the overaged condition. In the smaller-particle-size region, *i.e.* the underaged condition, prismatic dislocation loop formation determines the value of threshold stress and, in the large-particle-size region, *i.e.* the overaged condition, grain boundary dislocation climb determines the value of threshold stress. In the peak-aged condition, both prismatic dislocation loop formation and grain boundary edge dislocation climb control the creep process.

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