Letter

Effect of particle size on the creep rate of superalloy Inconel 718

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ABSTRACT

The effect of precipitate particle size on the steady state creep rate of commercial superalloy Inconel 718 has been investigated. It was observed that the steady state creep rate first decreased with increase in the γ'' precipitate size. However, after reaching a minimum value, it started to increase with increase in the γ'' precipitate size. The particle size for the minimum steady state creep rate was slightly smaller than that for the optimum room temperature strength, which has been attributed to the different deformation mechanisms of high temperature creep and the room temperature tensile tests.

Inconel 718, which is strengthened by 13 vol.% of coherent ordered disc-shaped body-centred tetragonal γ'' phase and 4 vol.% of spherical f.c.c. γ' precipitates, is one of the most widely used nickel-based superalloys for high temperature applications. It has an excellent yield strength as well as a reasonably good creep strength up to 923 K. The creep behaviour and deformation mechanism of this alloy in the temperature range 853–943 K and the stress range 620–865 MN m^{-2} have been studied recently [1, 2]. The experimental results revealed that (1) with increase in either the temperature or the applied stress the creep deformation mechanism changes from diffusional creep to dislocation power law creep, (2) the characteristics of diffusional creep in two-phase alloys are different from those observed in single-phase alloys and (3) the back stress in the present alloy, which opposes deformation during creep, is as high as 60%–95% of the applied stress and is influenced by the size of the γ'' and γ' precipitate particles. In this Letter the effect of particle size on the steady state creep rate of Inconel 718 at a temperature of 873 K and in the stress range 670–815 MN m^{-2} is presented.

The chemical composition of the material used in this study was 0.03 wt.% C, 19.0 wt.% Fe, 18.0 wt.% Cr, 0.5 wt.% Al, 1.0 wt.% Ti, 3.0 wt.% Mo, 5.0 wt.% Nb + Ta and the balance nickel. The creep specimens with a gauge length of 25.4 mm and a cross-section of 1.3 mm × 5.2 mm were machined from strips 1.3 mm thick cold rolled from the as-received sheets 3.4 mm thick. To obtain various γ'' and γ' particle sizes, the creep specimens were solution treated at 1323 K for 1 h, water quenched and then aged at 998 K for various periods of time from 1 to 100 h. A typical microstructure of an aged specimen is shown in Fig. 1. This figure is the dark field transmission elec-

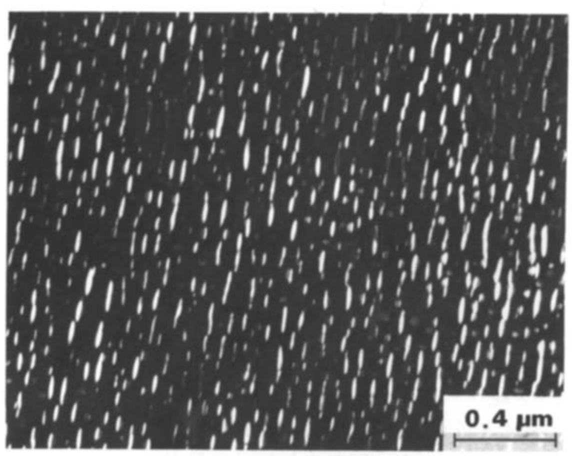

Fig. 1. Transmission electron micrograph of the microstructure, showing γ'' and γ' precipitates of Inconel 718 aged for 50 h at 998 K.

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tron micrograph of a specimen aged for 50 h at 998 K. In Fig. 1 the disc-shaped particles are γ'' phase and the spherical particles are γ' phase. The particle sizes of the γ'' and γ' precipitates were measured from dark field transmission electron micrographs as described earlier [3]. The creep tests were carried out in T-48 Avery–Denison constant-stress tensile creep machines at a temperature of 873 ± 2 K and applied stresses of 670, 695, 720, 765 and 815 MN m^{-2}. The experimental details of the tests have been reported earlier [1, 2]. The steady state creep rates were usually measured for 10–30 h after the creep rate had reached the steady state creep stage.

The dependence of the steady state creep rate $\dot{\epsilon}_s$ on the particle size $d_{\gamma''}$ at various applied stresses is graphically illustrated in Fig. 2. The results show that at all these stress levels the steady state creep rate first decreases with increasing particle size until a critical particle size is reached. It then increases with further increase in particle size. The results of the underaged specimens seem to be in general agreement with the models proposed by Ansell and Weertman [4] and by McLean [5], both of which suggest that $\dot{\epsilon}_s \propto d^{-2}$. In the overaged condition, however, results seem to follow the model of Grant [6] and McLean and Hale [7] who suggest that $\dot{\epsilon}_s \propto d^2$. It is observed in the present alloy that the particle sizes in specimens with an optimum creep resistance ($d_{\gamma''}$ = 23.2 nm and $d_{\gamma'}$ = 13.7 nm, for specimens aged for 10 h at 998 K) are slightly smaller than those observed in specimens exhibiting a maximum room temperature yield strength ($d_{\gamma''}$ = 27.0 nm and $d_{\gamma'}$ = 15.7 nm for specimens aged for 15 h at 998 K). This phenomenon is similar to that observed in Ni–Cr–Al–Ti alloys containing 10%–20% of γ' phase [8]. There are two possible reasons for this phenomenon.

Firstly, the deformation mechanism of high temperature creep is different from that of room temperature tensile deformation. It is known that, in the underaged condition, particles are cut by moving dislocations during room temperature deformation [9–11], while in high temperature creep, dislocations are believed to climb over the particles [2, 4]. It is also believed [12, 13] that the applied stress required for dislocations to cut or shear particles is higher than that required to climb over the particles, as shown in Fig. 3. In the overaged condition, during both room temperature deformation and high temperature creep, dislocations bow out between the particles and leave dislocation loops around the particles, forming dislocation pile-ups [2, 11]. However, in room temperature deformation, these dislocation pile-ups will result in a higher stress opposing dislocation motion but, during high temperature creep, the opposing stress due to pile-ups

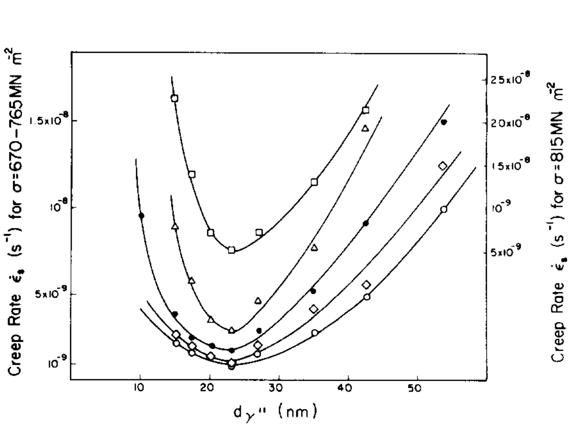

Fig. 2. Dependence of $\dot{\epsilon}_s$ on the particle size $d_{\gamma''}$ of the specimens aged for various periods of time at 998 K and tested at 873 K: ○, σ_a = 670 MN m^{-2}; ◇, σ_a = 695 MN m^{-2}; ●, σ_a = 720 MN m^{-2}; △, σ_a = 765 MN m^{-2}; □, σ_a = 815 MN m^{-2}.

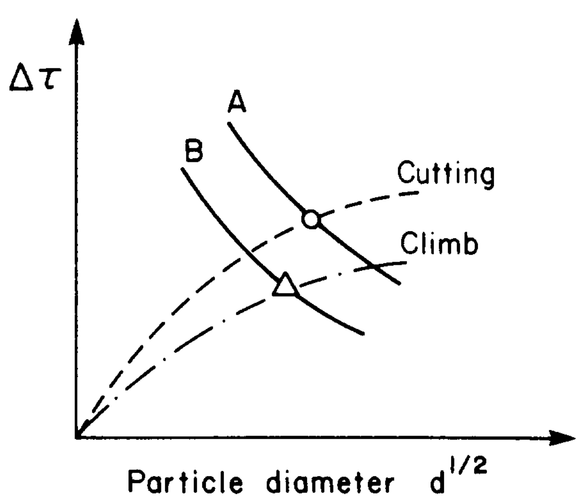

Fig. 3. Schematic aging-hardening curves, illustrating the particle size of the optimum room temperature yield strength (○) and the particle size of the optimum high temperature creep resistance (△): curve A, bowing-out at room temperature; curve B, bowing-out at a high temperature.

can be decreased by climb of the innermost loop and self-annihilation, permitting the next dislocation to bow out [4]. Therefore the applied stress required for dislocations to bow out between the particles during room temperature deformation will be higher than that during high temperature creep. Furthermore, higher thermal energy is available during high temperature creep than during room temperature deformation. This should assist the bowing out of dislocations. This is also shown in Fig. 3. It is known that the optimum particle size for room temperature yielding is obtained when the particle size increases to a dimension where the applied shear stress required for the dislocation to cut the particles is equal to that required for the dislocation to bow out between the particles. As a result, in the peak-aged condition, dislocations are seen to cut the particles as well as to bow out between them [14]. Therefore, this particle size during high temperature creep is generally slightly smaller than that where a maximum in yield strength is observed.

Secondly, although precipitates do not grow significantly when the specimen is held at 998 K for 20-30 h with no applied stress, an applied stress at this temperature may promote their growth. This has been observed in the specimen aged for 25 h at 998 K and crept for 222 h at 873 K and a stress of 720 MN m^{-2}. A particle size $d_{\gamma''}$ of 40 nm observed in this specimen is larger than the precipitate size $d_{\gamma''}$ of 32.4 nm observed in a specimen aged for 25 h at 998 K in the absence of any stress.

A discrepancy was also observed between the experimental and theoretical values of the exponent n in the equation $\dot{\epsilon}_s \propto d^n$, as listed in Table 1. It is observed that in the underaged region the value of n is between -1.9 and -2.5 for $\sigma_a \leqslant 765$ MN m^{-2} and -3.4 for $\sigma_a = 815$ MN m^{-2} instead of -2.0 as predicted by the expression $\dot{\epsilon}_s \propto d^{-2}$ for the underaged material [4, 5]. This rapid decrease in creep rate with increase in particle size may be a consequence not only of an increase in particle size but also of an increase in the volume fraction of precipitates. It has been observed that the volume fraction of precipitates increases with increasing aging time during the first 4 h of aging at 998 K and then remains constant [3]. Therefore, it is possible that a part of the decrease in the creep rate of specimens aged for up to 4 h is due to an increase in the volume fraction of precipitates. Secondly, the large discrepancy in the value of n when $\sigma_a = 815$ MN m^{-2}, *i.e.* a value of -3.4 instead of -2.0, may be explained as follows. In this situation the stress may be high enough for dislocations either to cut the precipitate particles or to bypass them by overriding their strain field. In the latter a small amount of localized bending by either cross-slip or climb causes the moving dislocation to glide continuously [5]. The creep rate in this case is faster than that predicted by Ansell and Weertman [4] where the rate is controlled by the climb of whole dislocation lines over the particles. The finer the particles, the easier the cutting or overriding process, and therefore the $\dot{\epsilon}_s \propto d^{-2}$ relationship breaks down. This may be the reason for the high creep rates observed in specimens with very fine particles. However, for the overaged condition, the value of n is between 2.4 and 2.5, which seems to be in a reasonably good agreement with the stored energy model of Grant [6] and McLean and Hale [7].

TABLE 1

Experimentally determined values of exponent n in the equation $\dot{\epsilon}_s \propto d^n$ for specimens tested at 873 K

Condition	$d_{\gamma''}$ (nm)	σ (MPa)	n
Underaged	<23	670	−1.9
Underaged	<23	695	−2.1
Underaged	<23	720	−2.0
Underaged	<23	746	−2.2
Underaged	<23	765	−2.5
Underaged	<23	815	−3.4
Overaged	>23	670	2.5
Overaged	>23	695	2.4
Overaged	>23	720	2.4
Overaged	>23	765	2.5
Overaged	>23	815	2.4

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