Materials Science and Engineering, A156 (1992) 53-61

Effect of cold and warm working on the microstructures and mechanical properties of alloy 718

M. C. Chaturvedi and Han Yafang
Mechanical Engineering Department, University of Manitoba, Winnipeg, Man. R3T 2N2 (Canada)

(Received August 27, 1990; in revised form January 17, 1992)

Abstract

This paper reports the influence of the amount and temperature of deformation on the mechanical properties, dislocation structure and coarsening of second phase particles in Alloy 718. The material was deformed 25% and 50% at room temperature, 723, 998 and 1173 K temperatures and then aged at 998 K for various lengths of time. The mechanical properties of the thermomechanically treated material were evaluated by room temperature tensile tests and elevated temperature creep tests. The dislocation substructures and coarsening behaviour of γ'' and γ' precipitates in both pre-deformed and non-predeformed material were studied by transmission electron microscopy. The experimental results show that the tensile yield strength and ultimate strength are greatly improved by cold and warm working prior to ageing but these increases are accompanied by a small loss in ductility. The results of creep tests in the temperature range 873-923 K, however, show that creep resistance is significantly improved by predeformation. A pre-ageing deformation of 25% at a temperature of 998 K followed by ageing for 15 h at 998 K is suggested to be the optimum thermomechanical treatment for Alloy 718.

1. Introduction

Thermomechanical treatment (TMT) of two phase nickel base superalloys can improve such mechanical properties as tensile yield strength, ultimate strength, low cycle and high cycle fatigue, and stress rupture life [1-6]. It has been reported that hot deformation in the appropriate recrystallization temperature region produces fine grain size and hence superplasticity, which introduces excellent formability in the material for the production of complex shaped components [1, 7, 8]. It has also been observed that cold working at room temperature usually greatly increases the yield and ultimate strengths with some loss of ductility or decreased resistance to high temperature creep [2, 6, 7]. With a proper amount of deformation, warm working near the polygonized temperature region can significantly improve the tensile yield strengths and other mechanical properties of materials with a minimum loss of ductility [1, 3-6]. The improvement in mechanical properties can be attributed mainly to the formation of dislocation substructures which act as obstacles to dislocation motion and also to the interaction between the second-phase particles and dislocations, since the second-phase particles can act as obstacles to dislocation motion as well as stabilizing the dislocation substructure. The effects of processing and deformation parameters of TMT on the mechanical properties and microstructures of Alloy 718 have been studied by several investigators [6, 7, 10]. The coarsening behaviour of γ'' and γ' phase during conventional ageing, strengthening mechanisms at RT, the effect of precipitate size on high temperature creep deformation of this alloy have been reported earlier [11-13]. In this communication the influence of the amount and temperature of deformation on mechanical properties, microstructures and growth characteristics of γ'' precipitates in Alloy 718 is presented.

2. Materials and experimental techniques

The starting material was commercially produced 3.2 mm thick sheets of Alloy 718 provided by International Nickel Company of Canada. The chemical composition of the alloy was 19Fe-18Cr-5(Nb+Ta)-0.5Al-1Ti-3Mo-0.3C-Ni Bal. Strips 20 mm wide cut from the as received materials were cold rolled to thicknesses of 2.8 and 1.8 mm. They were then solution treated at 1323 K for 1.5 h and quenched in water. The solution treated strips were rolled to 1.4 mm thickness at various deformation temperatures, including RT, 723 K, 998 K and 1173 K. Four and two rolling passes, with 5 min interpass anneals at deformation temperatures, were used to obtain a deformation of 50% and 25% reduction of area (RA) in 2.8 mm and

0921-5093/92/$5.00

1.4 mm thick strips respectively Tensile and creep specimens with a cross-section of 5.5 × 1.3 mm and a gauge length of 25.4 mm were machined from the deformed strips and then aged at 998 K for different periods from 2 to 100 h Tensile tests were carried out in a 1137 Instron universal testing machine at a strain rate of 3.3×10^{-4} s^{-1} Creep tests were carried out in a constant stress tensile creep machine in the stress range of 690 and 840 MN m^{-2} at 873 and 923 K.

Thin foil electron microscopy was used to examine the microstructures of the thermomechanically treated, and creep deformed materials Thin foils were prepared from discs of 3 mm diameter punched from tensile and creep specimens These discs were mechanically ground to a thickness of 0.15-0.20 mm then electropolished in a jet electropolishing unit using a 15% perchloric acid 85% methanol bath maintained at 223-233 K. To measure the dislocation cell size, five or more micrographs from different areas were taken and the average diameter obtained from the measurements of a minimum of 20 dislocation cells The particle sizes of the γ'' phase were measured from dark field micrographs taken with (001) reflection spots as described earlier [11]

3. Experimental results

3.1 Results of tensile tests

The grain size of the material solution treated at 1323 K for 1.5 h and used for various TMT was about 0.8 mm. Figures 1(a) and 1(b) show the variations in RT yield and ultimate strengths respectively *vs* ageing time for specimens deformed to various levels at different temperatures. Both the yield and ultimate strengths were increased significantly by TMT For example, after the same ageing condition of 15 h at 998 K, the yield strengths of the specimens deformed 50% at RT, 723, 998 and 1173 K were increased by 53%, 50%, 50% and 40% respectively when compared with the non-predeformed specimens For the 25% predeformed material the increases were 33%, 33%, 29% and 27% respectively The yield strengths of material predeformed at RT and 723 K decreased continuously with ageing time. However, the yield strengths first increased and then decreased for material which was predeformed at 998 K and 1173 K. The latter behaviour was similar to that of the non predeformed material, but the rates of increase and decrease in yield strength with ageing time were considerably more rapid in the non predeformed material Figures 2(a) and 2(b) are the plots of total plastic elongation and relative area under stress-strain curves, the latter being an indication of the toughness, for all nine treatments Although the yield and ultimate strengths of the

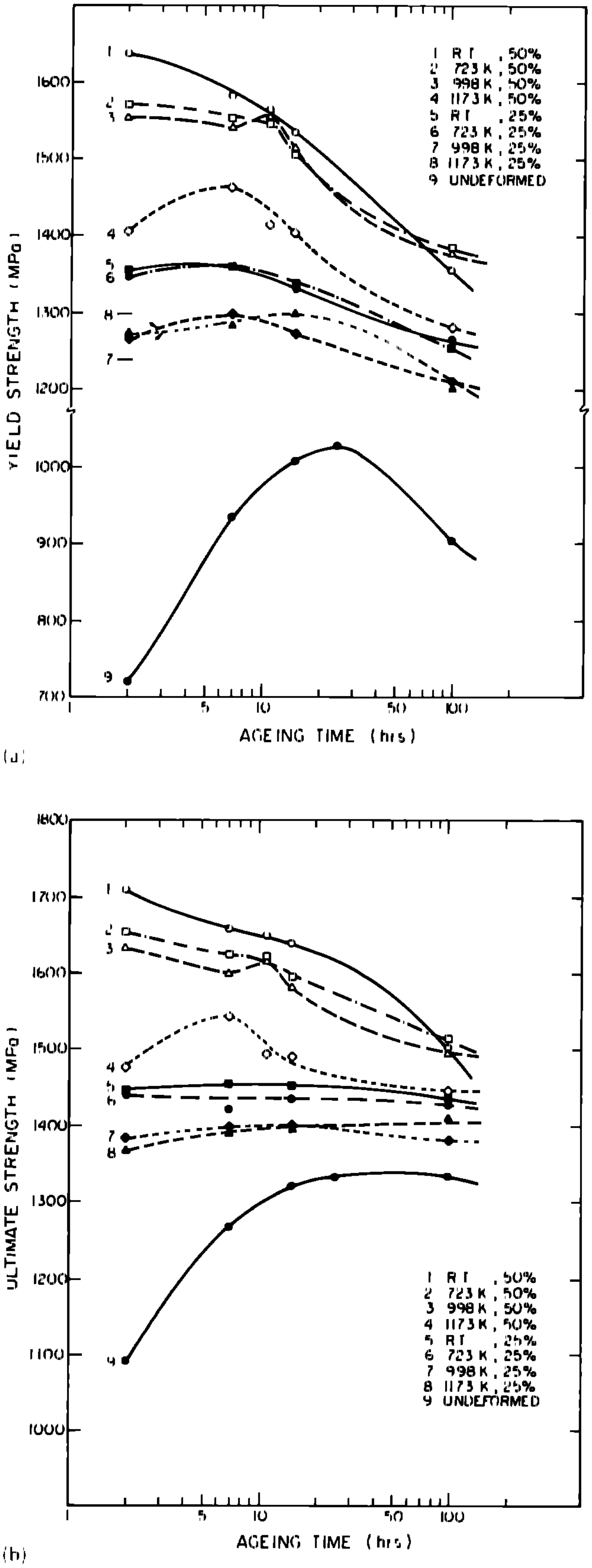
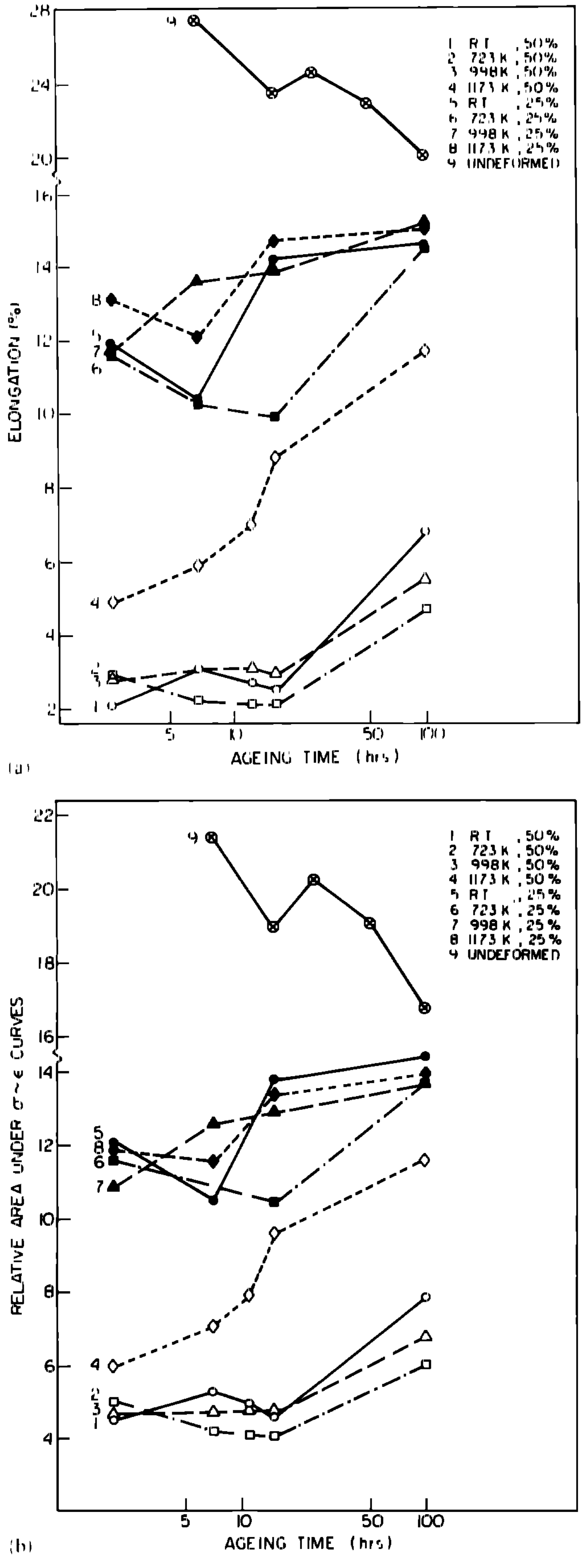


Fig 1 Results of room temperature tensile tests of deformed specimens aged at 998 K (a) yield strength, (b) ultimate strength



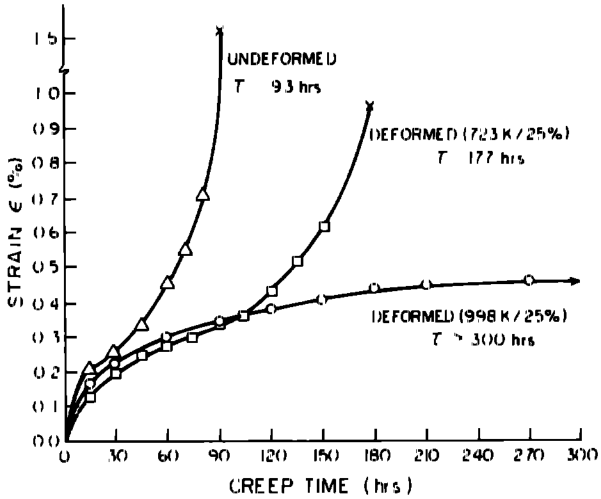


Fig. 3. Creep curves of deformed and undeformed specimens tested under the stress level of 690 MN m^{-2} at 923 K.

materials predeformed 50% have been greatly improved, the ductility and toughness (expressed by the area under stress-strain curve) are considerably reduced. The ductility and toughness of 25% pre-deformed material appear to be acceptable for many engineering applications.

3.2. *Results of creep tests*

The creep tests were carried out in the stress range of 690–840 MN m^{-2} at temperatures of 873 K and 923 K. Figure 3 shows the creep curves of 25% pre-deformed and non predeformed specimens aged for 15 h at 998 K and tested at a stress of 690 MN m^{-2} and a temperature of 923 K. The specimen prior deformed for 25% at 998 K exhibits the lowest creep rate and also a much longer stress-rupture life than those of the non predeformed specimen and the specimen predeformed 25% at 723 K. The steady state creep rates of the specimens predeformed 25% at 998 K and the non-predeformed specimens were determined in the stress range of 765 to 840 MN m^{-2} and at a temperature of 873 K. The results are shown in Table 1 and Fig. 4. It is observed that the creep rates of specimens predeformed 25% at 998 K are three times lower than those without prior deformation. The back stresses were determined by the consecutive stress reduction method [14]. The results, also given in Table 1, show that the back stresses of the pre-deformed specimens are about 10% higher than those of the non predeformed specimens. The values of the apparent and effective stress exponents, n_a and n_e, in

Fig. 2. Results of room temperature tensile tests of deformed specimens aged at 998 K: (a) total elongation; (b) relative area under stress-strain curve.

TABLE 1. Steady state creep rates $\dot{\varepsilon}_s$ and back stresses, σ_0 of deformed and undeformed alloy 718 in the stress range of 765-840 MN m^{-2} at 873 K

State of specimen	Creep rates ($\dot{\varepsilon}_s$) and back stress (σ_0)						
	σ_a (MN m^{-2})	765	790	815	840	n_a	n_e
1323 K/1.5 h 998 K/15 h	$\dot{\varepsilon}_s$ (s^{-1}) σ_0 (MN m^{-2})	2.78×10^{-9} 588	4.63×10^{-9}	6.06×10^{-9} 604	10.1×10^{-9}	13.5	5.1
1323 K/1.5 h deformed 25%/998 K 998 K/15 h	$\dot{\varepsilon}_s$ (s^{-1}) σ_0 (MN m^{-2})	8.10×10^{-9} 640	13.8×10^{-9}	17.8×10^{-9} 668	31.8×10^{-9}	12.7	4.4

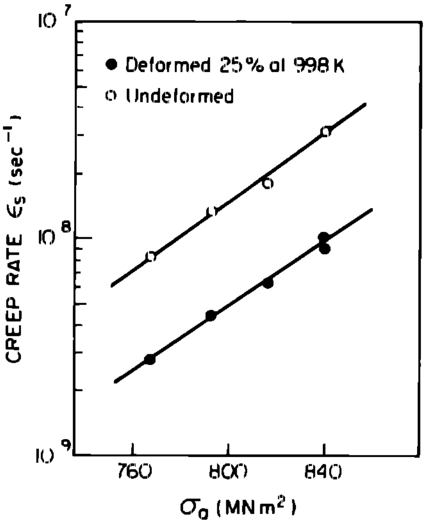


Fig. 4. Dependence of the steady state creep rate on the applied stress of deformed and undeformed specimens tested in the stress range of 765-840 MN m^{-2}.

the creep rate equations, $\dot{\varepsilon}_s \alpha \sigma_a^{n_a}$ and $\dot{\varepsilon}_s \alpha (\sigma_a - \sigma_0)^{n_e}$, were also determined. They were found to be n_a = 13.5 and 12.7, n_e = 5.1 and 4.4 for non predeformed and predeformed (25% at 998 K) specimens respectively, as shown in Table 1. This suggests that dislocation power law creep mechanism operates in both the non-predeformed and predeformed alloys in the stress range and temperatures used in this study.

3.3. *Dislocation structures of deformed materials*

The dislocation substructures of specimens after prior deformations of 25% RA and 50% RA at RT, 723, 998 and 1173 K were examined by thin foil electron microscopy. The dislocation cells had well defined walls and were nearly free from interior dislocations in all eight specimens. This suggests that deformation levels to stage II or III had been reached [15]. Typical dislocation substructures are illustrated in Fig. 5. Figures 5(a), 5(b), 5(c) and 5(d) show the dislocation cells in specimens with a prior deformation of 50% at temperatures of RT, 723 K, 998 K and 1173 K respectively. Figures 5(e) and 5(f) are the dislocation substructures in specimens deformed 25% at 998 K and 1173 K respectively. At a given deformation temperature, the size of dislocation cells decreases with increasing deformation and increases with increasing deformation temperature for a constant amount of deformation. When either the deformation temperature was decreased or the degree of deformation was increased, the walls of dislocation cells became thicker. The dislocation cell size in the predeformed specimen was measured and is listed in Table 2. The dependence of dislocation cell size on deformation temperature is also plotted in Fig. 6, which shows an increase in the cell size with deformation temperature.

The dislocation structures of crept specimens were also examined and examples are shown in Fig. 7. The dislocation cell walls formed during predeformation of 25% at 998 K were nearly absent after the creep test. Relatively homogeneously distributed dislocation networks were present in the specimens tested for 130 h at 815 MN m^{-2}/923 K (Fig. 7(a)) and for 270 h at 720 MN m^{-2}/873 K (Fig. 7(b)).

3.4. *Precipitation and growth of γ″ and γ′ phases during ageing of TMT materials*

As with non predeformed Alloy 718 [11], two kinds of coherent ordered precipitates formed during the early stage of ageing of the predeformed materials, i.e. disc-shaped b.c.t. γ″ phase and spherical f.c.c. γ′ phase. The morphology and size of γ″ and γ′ phases are shown in Figs. 8(a) and 8(b) which are the bright- and dark-field micrographs respectively of a specimen predeformed 25% at 998 K and aged at 998 K for 100 h. The particle sizes of γ″ precipitates in deformed specimens aged at 998 K for 100 h were measured from the dark field micrographs by taking the average value of the diameter of 50 γ″ particles. The results are listed in Table 2. For comparison, the γ″ particle sizes in the specimens without TMT were also measured and the results included in Table 2. These results suggest that the size of γ″ precipitates increases with deformation temperature and decreases with an increase in the prior strain value. The results also show that γ″ particle sizes in the non predeformed specimen are

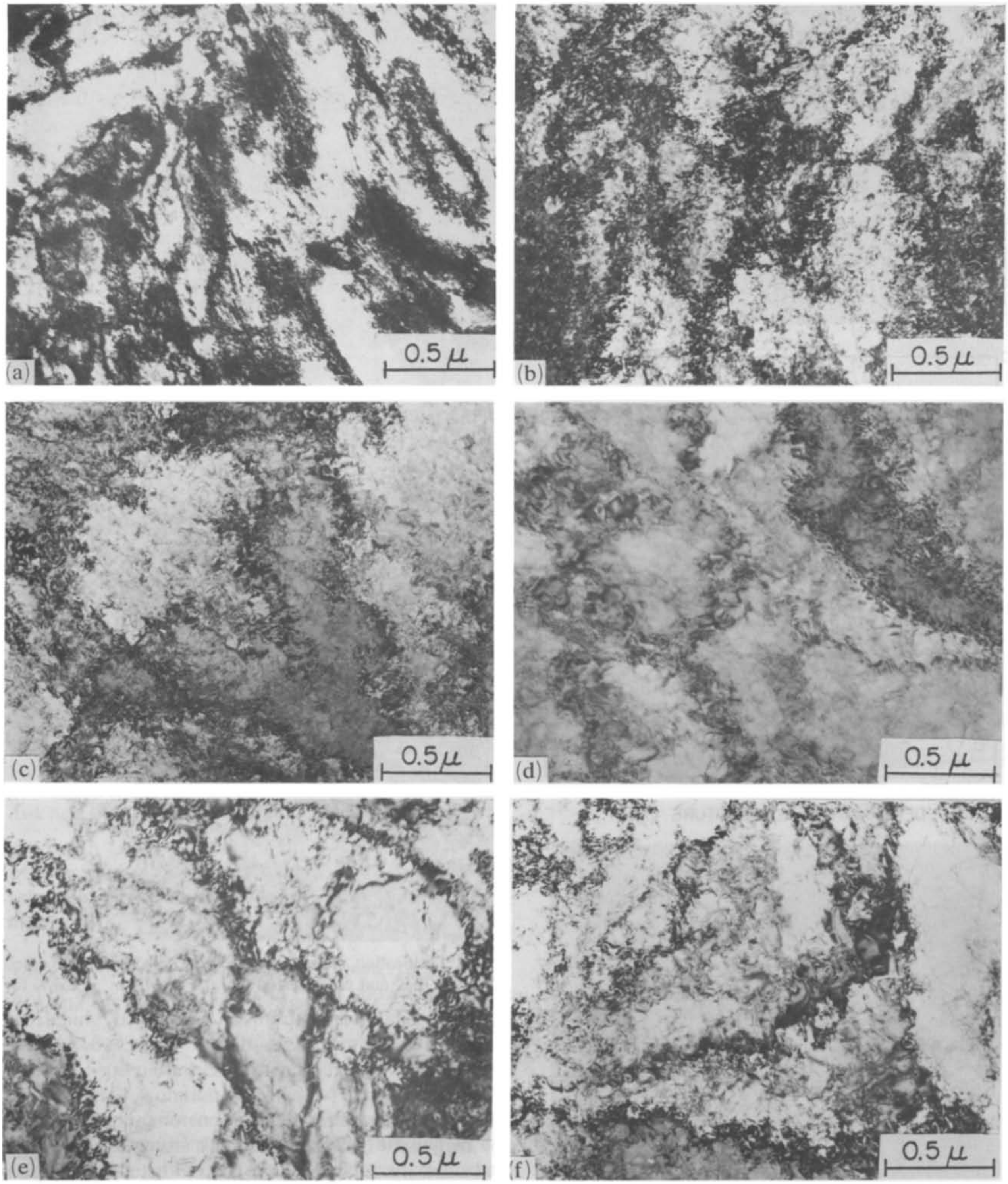


Fig. 5. Dislocation substructures of deformed specimens: (a) 50% r.t.; (b) 50%/723 K; (c) 50%/998 K; (d) 50%/1173 K; (e) 25%/998 K; (f) 25%/1173 K.

TABLE 2. Dislocation cell size and diameter of disc shaped, γ'' phase d γ''

Deformation		Dislocation cell size (μm) No ageing	d γ'' (nm) 998 K/15 h	d γ'' (nm) 998 K/100 h
Temperature (K)	RA			
RT	50%	0.43	20.8	29.8
723	50%	0.45	22.1	33.0
998	50%	0.53	22.3	35.9
1173	50%	0.57	23.4	50.6
RT	25%	0.60	–	40.9
723	25%	0.61	–	43.5
998	25%	0.73	–	45.5
1173	25%	0.82	–	53.6
Undeformed		–	27.2	58.1

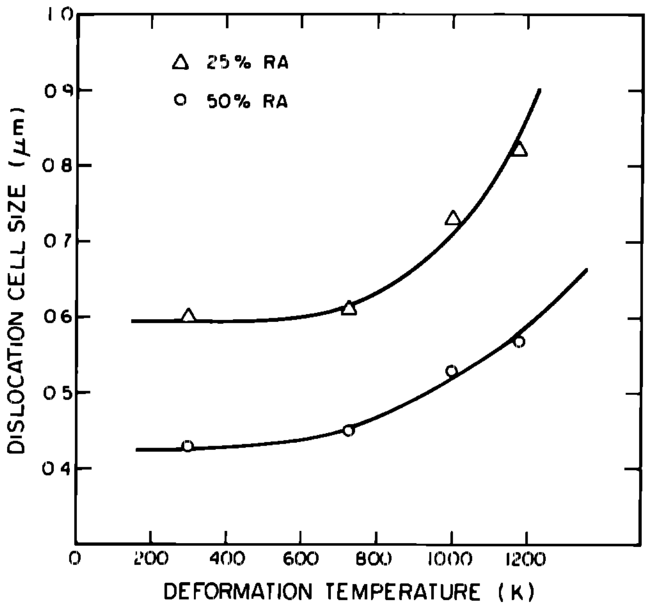


Fig. 6. Dependence of dislocation cell size on deformation amount and temperature.

larger than those observed in the predeformed specimens for the same period of ageing. This suggests that the existence of a dislocation substructure after TM treatment may refine the distribution of γ'' and γ' precipitates, as shown in Fig. 9.

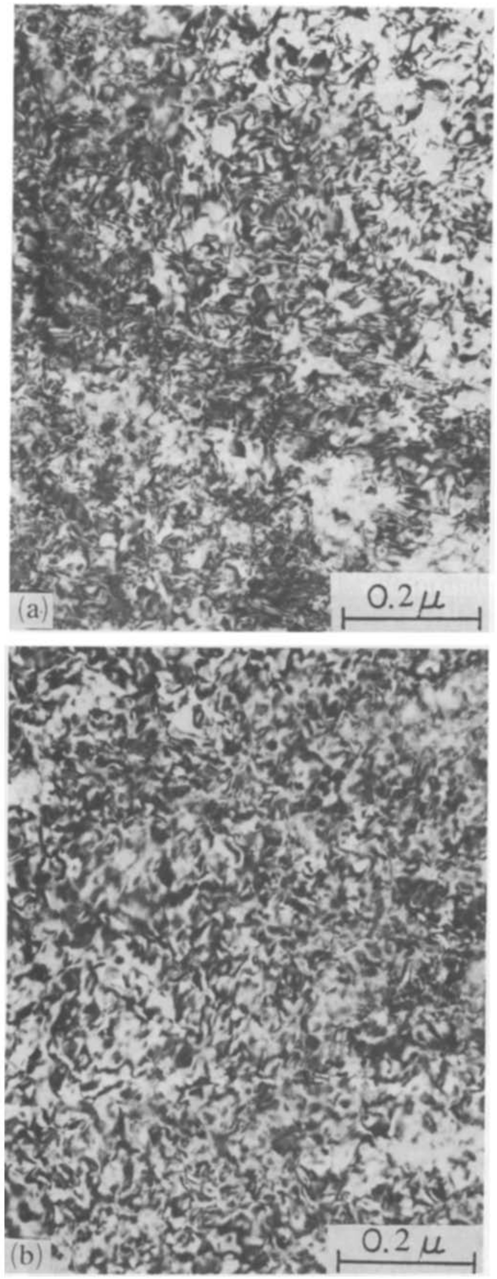


Fig. 7. Dislocation substructures of the specimens deformed 25% at 998 K and creep tested at (a) 815 MN m^{-2}/923 K for 130 h, (b) 815 MN m^{-2}/873 K for 30 h and 720 MN m^{-2}/873 K for 270 h.

4. Discussion

4.1 Effect of deformation amount and temperature on yield strength

The experimental results show that the yield and ultimate strengths of cold and warm worked Alloy 718 increased with the amount of predeformation. This increase in the strength of predeformed specimens can be attributed to an increase in dislocation density in the material. The greater the amount of predeformation, the higher the yield and ultimate strengths. The yield and ultimate strengths were also affected by the predeformation temperature *i.e.* a higher predeformation temperature resulted in a lower yield strength. This may be due to the decrease in dislocation density and the increase in dislocation cell size with the increase in predeformation temperature as was observed in this investigation. This may be attributed to the recovery

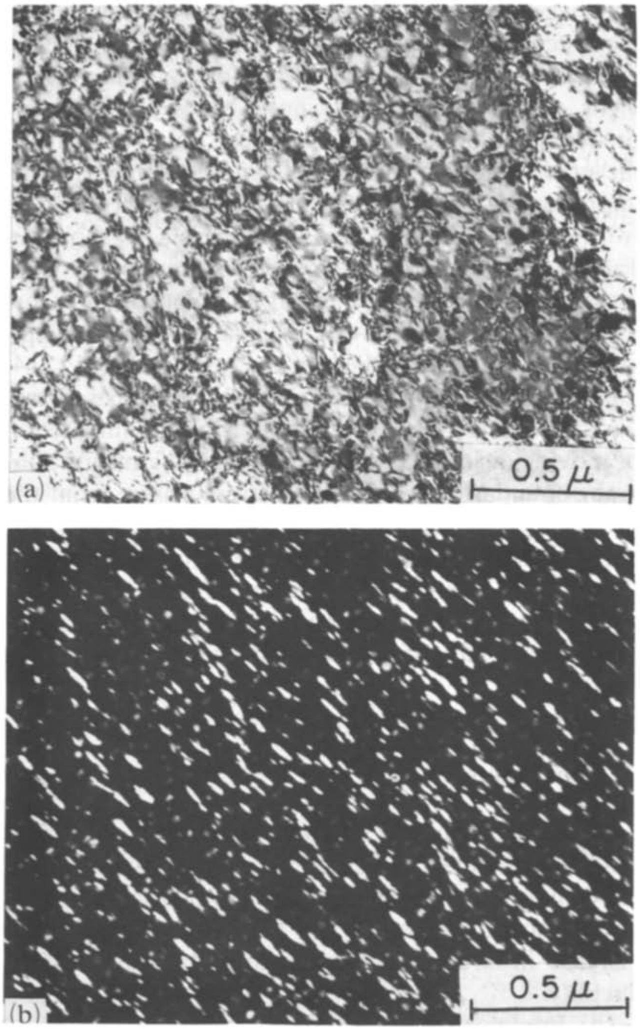


Fig. 8. Precipitates of γ'' and γ' phases in specimen deformed 25% at 998 K and aged for 100 h at 998 K: (a) bright field; (b) dark field taken with (100) reflection spot.

process during the interpass anneals of 5 min at the relatively higher predeformation temperatures of 998 and 1173 K.

4.2. Ageing behaviour of the predeformed Alloy 718

The experimental results in Fig. 1 show that the age hardening behaviour of predeformed specimens of Alloy 718 is somewhat different from that observed in the non-predeformed material. For the non-predeformed solution treated specimens aged at 998 K, the yield strength first increased rapidly with ageing time until a maximum yield strength was reached after about 15-25 h, and then decreased rapidly with further increase in ageing time. This was also found in a previous study [12]. However, for the 25% predeformed specimens and the specimens predeformed 50% at the temperatures of 998 K and 1173 K, the tensile strength vs. ageing time curves were flatter than those of non-predeformed specimens

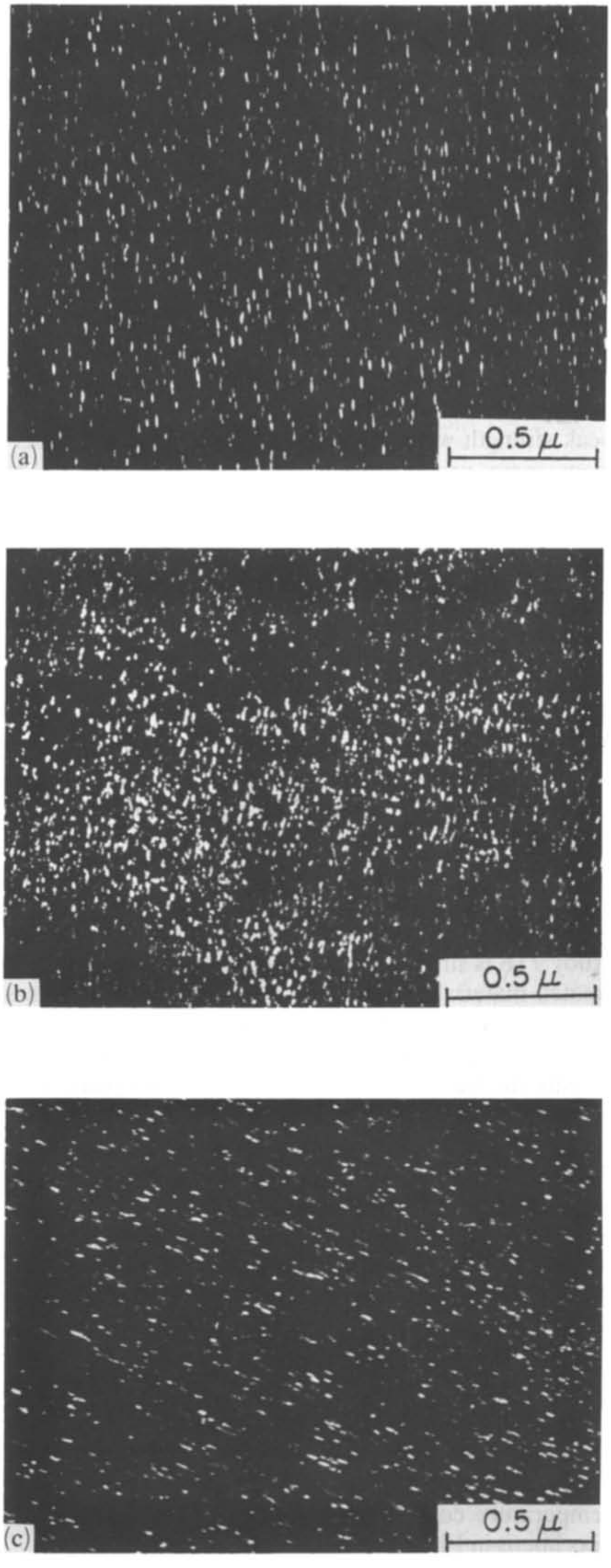


Fig. 9. Distribution and size of second phase precipitates: (a) 1323 K/1.5 h + 998 K/15 h; (b) 1323 K/1.5 h + 25%/998 K deformation + 998 K/15 h; (c) 1323 K/1.5 h + 50%/1173 K deformation + 998 K/15 h.

although the peak strengths in these ageing curves still appeared, as shown in Figs. 1(a) and 1(b). For the 50% predeformed material, the yield strengths decreased with ageing time when the specimens were pre-deformed at the temperature of r t. and 723 K. The results in Fig 1 show that a peak in the ageing curve was also present when the specimens were pre-deformed at the relatively higher temperatures of 998 K and 1173 K, as with the age hardening behaviour of non-predeformed material. The time to acquire peak strength for predeformed specimens were usually shorter than that for the non predeformed specimens, depending on the amount and temperature of predeformation. For example, the times to attain peak strength were about 7 and 10 h when the specimens were predeformed 50% at 1173 and 998 K respectively These different age hardening and softening behaviours of predeformed specimens are considered to be related to two competing processes occurring simultaneously during ageing before reaching the peak aged stage One is the softening process due to the growth of dislocation or cell size and the decrease in dislocation density. The second is the age hardening process due to the precipitation of the second phase particles of γ'' and γ' phases. These two competing processes result in relatively flat curves of yield strength or ultimate strength against ageing time in the early stage of ageing for the predeformed specimens

The experimental results in Table 2 show that the size of the γ'' precipitates in cold and warm worked Alloy 718 is smaller than that observed in the solution treated material. The size of the γ'' particles increases with increasing predeformation temperature and decreasing prestrain, *i e* the higher the dislocation density the finer the distribution of second-phase particles. This suggests that the pre existing dislocations refine the distribution of γ' and γ'' precipitates in Alloy 718, as shown in Fig 9. It appears that a significant interaction occurs between dislocations and second phase particles in the present alloy Dislocations seem to act as nucleating sites for fresh precipitation, producing a finer distribution of the second-phase particles, as shown in Fig. 9 In addition, the second-phase particles seem to act as the pinning points for subsequent dislocation movement and stabilize the dislocation substructure. This kind of dislocation and second phase interaction retards overageing of this alloy, making ageing curves relatively flat after peak temperature comparison to those of the undeformed specimens in Fig. 1

4.3. Improvement of creep resistance by predeformation

The results in Table 1 and Figs 3 and 4 show that the creep resistance of Alloy 718 is significantly improved by predeformation The stress rupture life of the specimens predeformed 25% at 998 K was at least three times longer than that of non-predeformed specimens, and the steady-state creep rates of specimens predeformed at 25% at 998 K were only one-third of the creep rates of non predeformed specimens. This improvement of creep resistance may be attributed to two factors. (i) an increase in back stress, as shown in Table 1, due to the formation of dislocation substructures; (ii) a refinement of second-phase precipitates due to the interaction between dislocations and precipitates, as described earlier It has been proposed [14] that the back stress in two phase alloys consists of two parts: (i) the back stress resulting from the interaction of moving dislocation with other dislocations and sub grain boundaries; (ii) the back stress arising from the presence of second phase particles For both cases, the back stress in predeformed materials will be higher than that in non-predeformed specimens. The higher back stress in the former case results in a decrease in the effective stress for creep and hence a lower creep rate.

5. Conclusions

(i) Tensile yield and ultimate strengths of Alloy 718 can be significantly increased by cold and warm working with a predeformation of 25% to 50% at deformation temperatures of r.t. to 1173 K However, the increase in strength is accompanied by some loss in ductility

(ii) Predeformation changes the age hardening behaviour of Alloy 718, the interaction between pre-existing dislocation and second phase particles not only stabilizes the dislocation substructure but also refines the distribution of second-phase particles and hence retards overageing of the alloy

(iii) Creep resistance of Alloy 718 can be significantly improved by predeformation of 25% at 998 K

Acknowledgments

The authors would like to thank the Natural Sciences and Engineering Research Council of Canada for financial support and the International Nickel Company of Canada for supplying the material, and the Beijing Institute of Aeronautical Materials for giving Y Han a leave of absence to work on this project.

References

1 B. H. Kear, J M. Oblak and W A Owczarski *J Met*, *24* (1972) 25

2 C. P. Blankenship, *Aerospace Structure Materials*, NASA, 1969, p. 91.
3 R. K. Hotzler, R. J. Maciag, G. J. Fischer and E. Troc, *Metall. Trans.*, *1* (1970) 963.
4 J. M. Oblak and W. A. Owczarski, *Metall. Trans.*, *3* (1972) 617.
5 M. N. Menon and W. H. Reimann, *Metall. Trans. A*, *6* (1975) 1075.
6 M. A. Meyers and R. N. Orava, *Metall. Trans. A*, *7* (1976) 179.
7 R. S. Cremisio, H. M. Butler and J. F. Radavich, *J. Met.*, *21* (111) (1969) 55.
8 J. B. Moore and R. L. Athey, US Patent 3519, p. 503 (quoted from ref. 1).
9 J. B. Moore and R. L. Athey, *Metal Working News*, *10* (1969) 10, *Design News*, January 5, 1970, p. 17 (from ref. 2).
10 Z. Jin, *M.Sc. Thesis*, Institute of Aeronautical Materials, Beijing, China, 1985.
11 Y. Han, P. Deb and M. C. Chaturvedi, *Met. Sci.*, *16* (1982) 555.
12 M. C. Chaturvedi and Y. Han, *Met. Sci.*, *17* (1983) 145.
13 Y. Han and M. C. Chaturvedi, *Mater. Sci. Eng.*, *87* (1987) 25.
14 Y. Han and M. C. Chaturvedi, *Mater. Sci. Eng.*, *85* (1987) 59.
15 N. Hansen and Kuhlmann-Wilsdorf, *Mater. Sci. Eng.*, *8* (1986) 141.
16 Y. Han and M. C. Chaturvedi, *Mater. Sci. Eng.*, *89* (1987) L7.