JMEPEG (2000) 9:324-329

Tensile Strengthening in the Nickel-Base Superalloy IN738LC

E. Balikci, R.A. Mirshams, and A. Raman

(Submitted 1 June 1999; in revised form 3 February 2000)

The tensile properties of superalloy IN738LC with different precipitate microstructures are evaluated at room temperature, 650 °C, 750 °C, and 85 °C at two different strain rates. The properties can be presented in two groups based on the comparable closeness of the values obtained—those of microstructures C and M, with coarse and medium size precipitates, and those of microstructures F and D, with fine and duplex size (medium + fine) precipitates. Preferred orientations, lattice parameters, and metallography are used to characterize the microstructure and tensile testing to determine the yield strength, tensile strength, and strain hardening coefficients. An anomalous increase in yield strength is observed, which occurs at temperatures about 100 °C higher with higher strain rate than with lower strain rate applied. The experimental results show that the yield strength is influenced by preferred orientations and precipitate size, while the tensile strength is effected by the size and morphology of precipitates.

Keywords alloy IN738LC, mechanical properties, tensile stress, yield stress

1. Introduction

IN738LC (Inconel 788 Low Carbon) is used in gas turbine engines at high temperatures. The unique mechanical properties required in the superalloy IN738LC are realized by having an fcc, Ni-base solid solution matrix that is hardened by suitable solutes and fine precipitates. Like many Ni-base superalloys, IN738LC is strengthened by the γ' precipitates of $Ni_3(Al, Ti)$ basic composition and $L1_2$ crystal structure. Westbrook[1] first reported on an anomalous rise in yield stress with temperature in superalloys. Studies of the single-phase Ni_3Al intermetallic alloy have also shown an analogous increase in the yield strength with increasing temperature with a maximum in the range of 600 to 800 °C.[2] The effect of morphology and volume fraction of γ' precipitates in the γ solid-solution matrix has already been reported in the literature. Anton[3] and Ardell[4] reviewed the interaction between γ'-Ni_3Al precipitates and γ solid-solution matrix in nickel-base superalloys. Anton pointed out the importance of the vast interfacial area between the γ solid-solution matrix and the ordered γ' precipitates during deformation. In addition, he explained the high-temperature strength of superalloys through intimate association between the two *via* their coherent nature and immobile dislocation as modeled by Kear-Wilsdorf lock.

Anomalous strengthening with the γ' precipitates is observed in IN738LC, but it is not as pronounced as in the case of pure Ni_3Al. However, the volume fraction of γ' precipitates and their morphology, *i.e.*, coarsening, γ/γ' mismatch, duplex microstructure, and stress coarsening, are significant on the anomalous behavior in the yield stress. Bettge *et al.*[5] conducted a series of experiments on a cast IN738LC containing 43% γ' with morphologies in the dentritic cores. They reported the effect of strain rate on a strong appearance of a minimum yield stress in an intermediate-temperature range around 45 °C. The yield flow stress dependency of strain rate in IN738LC is also reported and discussed in the literature.[3] The cube cross-slip mechanism and the formation of Kear-Wilsdorf locks have been used to model this intermediate-temperature and high-temperature thermally activated strengthening, while the mechanisms behind the low and intermediate strength lie in the classical dispersion strengthening theory.

Several new microstructures have been developed in IN738LC through the recent microstructure evolution analysis using various aging treatment procedures.[6] Fracture behavior and ductility of IN738LC have been investigated and reported for four different size precipitate microstructures recently.[7] The results indicate that the fracture feature correlates very well to ductility and fracture toughness of the specimens. In particular, cuboidal and medium size precipitate microstructures show better ductility and a dimple mode of fracture, while fine and duplex size precipitate microstructures exhibit lower ductility and undergo cleavage fracture. In the current paper, there are results of an investigation on the correlation of tensile strengthening in the Ni-base superalloy IN738LC with different precipitate microstructures at different high temperatures.

2. Experimental Procedure

2.1 Material

The material used in this research was produced by investment casting. The as-received cast alloy bars (15 mm diameter in 110 mm length) were hot isostatically pressed at 1170 °C/102 MPa for 2 h and solution treated at 1120 °C for 2 h and then aged at 843 °C for 24 h. All processing on the as-received stock was performed in a controlled atmosphere, either argon or vacuum at the Howmet Corporation facility (Whitehall, Mi). The chemical composition of the as-received material is given in Table 1.

E. Balikci and **A. Raman,** Mechanical Engineering Department, Louisiana State University, Baton Rouge, LA 70803; and **R. A. Mirshams,** Mechanical Engineering Department, Southern University, Baton Rouge, LA 70813. Contact e-mail:mirshams@engr.subr.edu.

Table 1 Chemical composition of the as-received IN 738LC (wt.%)

Ni	Cr	Co	Mo	W	Ta	Nb	Al	Ti	B	Zr	C
Balance	15.7–16.3	8–9	1.5–2	2.4–2.8	1.5–2	0.6–1.1	3.2–3.7	3.2–3.7	0.007–0.012	0.03–0.08	0.09–0.13

2.2 Microstructure Development and Metallography

Different precipitate microstructures were developed using the heat treatment schedules described elsewhere.[6] Briefly, heat treatment samples were cut from the cylindrical bars, wrapped individually with stainless steel foil, and then sealed in silica tubes under vacuum. The as-received material was solution treated at 1200 °C for 4 h and then water quenched to room temperature. The aging processes were carried out in the range of 1120 to 1200 °C for different periods followed by cooling in a furnace or water quenching. For microstructural studies, the conventional metallographic procedures were used for the specimen preparation. The etching solution composition was 33% HNO_3 + 33% acetic acid + 1% HF + 33% $H_2$0. A Hitachi (Hitachi, Ltd., Japan) S-2460N type scanning electron microscope was used in the backscatter mode to characterize the microstructure and morphology of the γ' precipitates. Among the precipitate micro-structures, four of them were tested for their tensile properties. These microstructures consisted of (1) fine size (F; ~70 nm size), (2) medium size, unimodal cuboidal (M; ~450 nm size), (3) coarse size, unimodal cuboidal (C; ~700 nm size), and (4) duplex size (D; ~50 and ~450 nm size) precipitates. The precipitate sizes are averages of the linear precipitate sizes extracted from the respective micrographs with 10.16 12.70 cm areas. These precipitated microstructures are shown in Fig. 1.

2.3 Tensile Testing

For the tensile testing and corresponding specimen preparation, ASTM E8 and ASTM E21 procedures were adopted. Round specimens with threaded ends were machined with the gage length of 3.18 cm and gage diameter of 0.51 cm. The tensile test specimens were heat treated exactly with the same procedure explained earlier to obtain the four different microstructures. An MTS 810 Test Star Material Testing System was used for tensile testing. An ATS three-zone high-temperature split furnace, equipped with a ceramic-probe high-temperature extensometer, was attached to the system, enabling high-temperature testing. The tensile tests were carried out at room temperature, 650 °C, 750 °C, and 850 °C. There was no controlled atmosphere in the furnace. Tests were carried out at two different strain rates (10^{-3} and 5×10^{-5} s^{-1}), and two specimens were tested for each microstructure, temperature, and strain rate.

2.4 X-ray Diffraction

About 1 mm thick rounds of samples with these microstructures were cut from the 15 mm diameter bar using a diamond cut-off wheel in the presence of a continuously flowing coolant. In general, the circular plates were cut parallel to the radial plane of the heat-treated specimens, corresponding to the radial section of the as-received bars. Heat treatment was undertaken after cutting the specimens in a few cases. In most cases, however, heat-treated specimens were subsequently cut and analyzed. The specimens were then subjected to X-ray diffraction (XRD) analysis using Cu K_α X-rays in a computer-controlled Siemens (Siemens Electrical Equipment, Toronto, Canada) D5000 diffractometer, equipped with a Kevex (Kevex X-Ray, Inc., CA) Psi Peltier-cooled silicon detector. The scan range was between 6 and 100° of 2θ in all cases. Preferred orientation information and lattice parameters of phases were extracted from the XRD patterns of samples from five different batches. More detailed information is provided in Ref 11.

3. Results

The tensile test results obtained for the different microstructures at different high temperatures are illustrated in Fig. 2 to 4, and the observations are given below for each microstructure separately.

3.1 Fine Size Precipitate Microstructure

This microstructure consists of unimodal, fine precipitates of size about 70 nm, as can be seen in Fig. 1(a). Specimens with such fine precipitates show the highest yield strength among the tested microstructures from room temperature up to 750 °C at both the strain rates applied. The variations of 0.2% offset yield strength values with temperature for this microstructure are given in Fig. 2(a) and (b). At the higher strain rate of 10^{-3} s^{-1}, the yield strength steadily decreases with increasing temperature up to 750 °C, but at 850 °C, a sharp decrease in the yield strength is observed. At the lower strain rate of 5×10^{-5} s^{-1}, the yield strength drops from the room temperature value to a lower one at 650 °C, and subsequently shows an anomalous increase up to 750 °C. Beyond this temperature, however, the yield strength drops drastically. The tensile strength of this microstructure decreases generally, though only by a relatively small amount from room temperature to 650 °C at the higher strain rate, but a sharp decrease is observed at and above 750 °C, as seen in Fig. 3. Strain hardening is generally small, but increases slightly at 750 °C; the values of strain hardening index ($n = d\sigma/d\varepsilon$) are nearly the same as at room temperature and 650 °C, and the lowest value (zero) is at 850 °C after the tests with both the strain rates (Fig. 4).

3.2 Medium Size Precipitate Microstructure

This microstructure, shown in Fig. 1(b), has unimodal cuboidal precipitates with a size of about 450 nm. Its yield strength decreases continuously up to 750 °C. In addition, a relatively sharp drop is observed in the range 750 to 850 °C with the lower strain rate. The tests with the higher strain rate produce nearly similar results, but with a slight, anomalous increase in the yield strength in the range 750 to 850 °C. The tensile strength increases slightly up to 650 °C at the lower strain rate. However, there are continuous decreases in tensile strength beyond this temperature at both strain rates. The strain hardening index in-

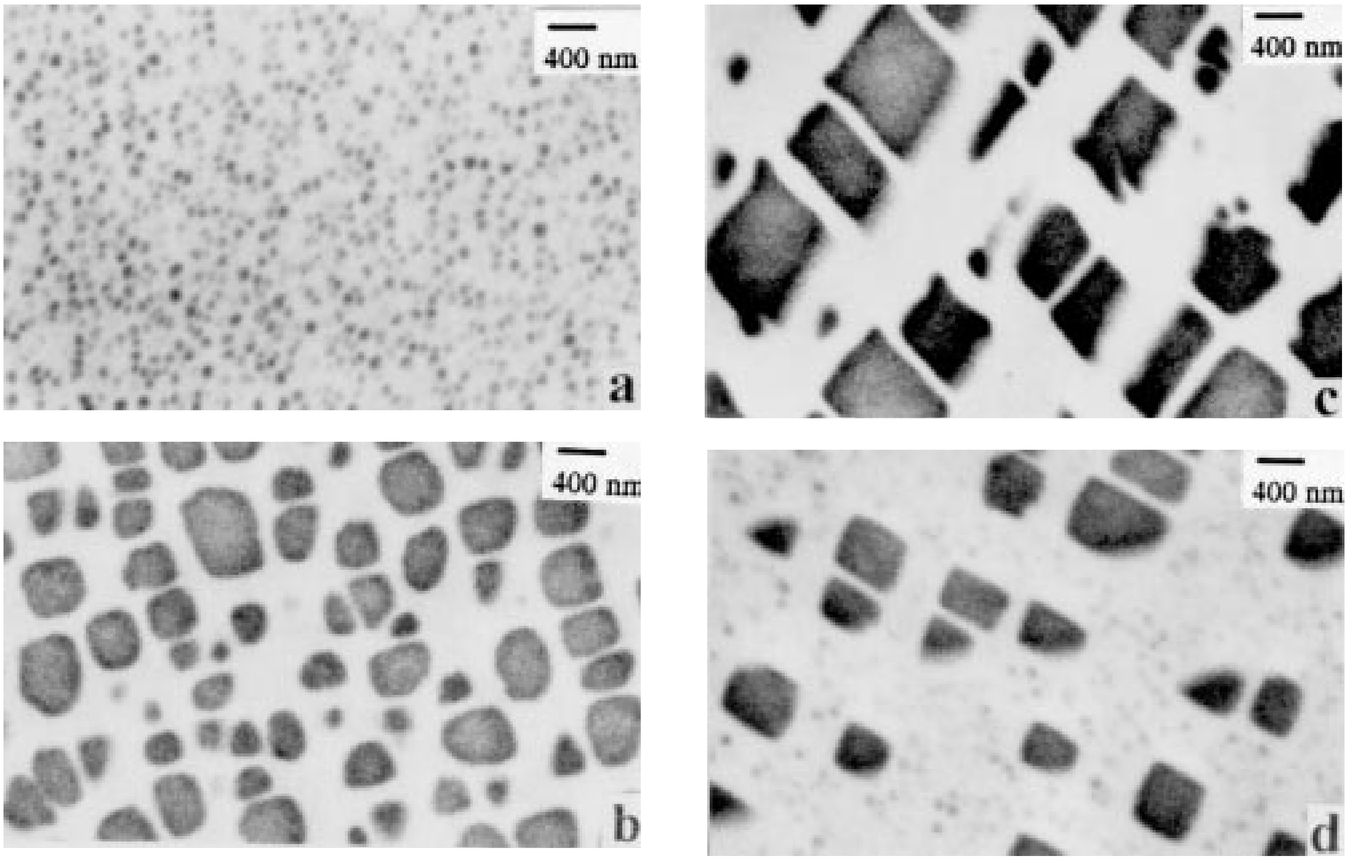


Fig. 1 Various precipitate (ppt) microstructures of IN738LC used in the tensile testing: (**a**) fine size ppts (F; ~70 nm), (**b**) medium size ppts (M; ~450 nm), (**c**) coarse size ppts (C; ~700 nm), and (**d**) duplex size ppts (D; ~50 and 450 nm)

creases by about 50% at 650 °C from its room temperature value; above this temperature, it decreases again at both strain rates.

3.3 Coarse Size Precipitate Microstructure

This microstructure has unimodal cuboidal precipitates of size about 700 nm (Fig. 1c). This is the coarsest precipitate size that was studied in this program. The yield strength of this microstructure exhibits similar behavior to the medium size precipitate microstructure at both strain rates. The anomalous strengthening peak is seen at 750 °C with the lower strain rate, and it shifts to 850 °C with the higher strain rate. The tensile strength increases very slightly at 650 °C, and it decreases beyond this temperature with both the strain rates. Maximum observed strain hardening among all microstructures is seen at 650 °C for this microstructure with both the strain rates (Fig. 4).

3.4 Duplex-Size Precipitate Microstructure

Two distinctly different precipitate sizes, one fine (~50 nm) and the other medium size (~450 nm), constitute this microstructure (Fig. 1d). The yield strength behavior of this microstructure is very similar to those of the microstructures M and C, though the value is distinctly higher and closer to that of material F. This microstructure has a distinctly higher tensile strength among all of those at Room Temperature after the tests at the higher strain rate, and it also shows a higher tensile strength at 850 °C at both the strain rates. The tensile strength decreases steadily, however, with increasing temperature at both the strain rates. The strain hardening index is generally small as it is for F; it seems to rise slightly up to 650 °C, but later it drops to zero at 850 °C with the lower strain rate (Fig. 4). A continuous decrease in the value of the strain hardening index is observed while using the higher strain rate from room temperature to 750 °C, and there is a steep drop in strain hardening to a nearly zero value in the range 750 to 850 °C.

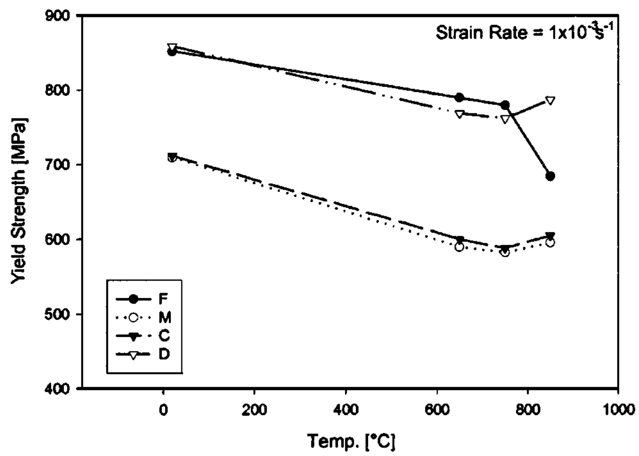
3.5 Microstructure and Preferred Orientations

Table 2 summarizes the results of metallographic microstructure characteristics and preferred orientation analysis of IN738 LC bars used in this investigation. More details are given in Ref 11.

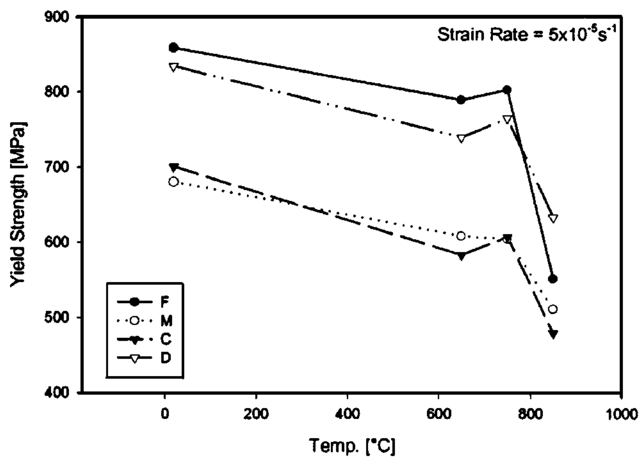
Calculation of the lattice parameters of the matrix and precipitates on the basis of experimental results[11] indicates that the relative misfit between the matrix fcc γ phase and the fcc γ' precipitate phase is rather small and is in the range 0.17 to 0.36%. It is concluded that the misfit is nearly zero in the superalloy used in this investigation. This is not unusual and such results have been reported in the literature for some other superalloys.[12]

Table 2 Summary of precipitate size and preferred orientation

Microstructure	Precipitate size (nm) [±12%]	Preferred orientation
C	700	{100}
M	450	{100}
F	70	{110}
D	50 and 450	{111} matrix {100} ppt



(a)



(b)

Fig. 2 Variation of 0.2% offset yield strength vs temperature at (**a**) high strain rate (10^{-3} s^{-1}) and (**b**) low strain rate (5×10^{-5} s^{-1})

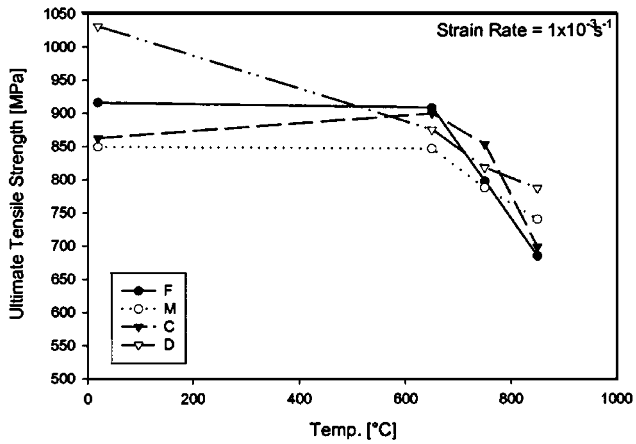
4. Discussion

4.1 Yield Behavior

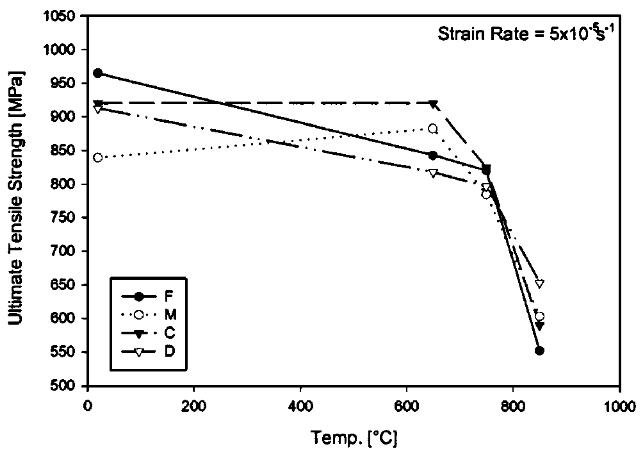
Changes in the precipitate size create a great influence on the γ/γ' interfacial area and therefore affect dislocation motion during deformation. The yield behavior of the microstructures tested in this work can be studied in two groups: the behavior of F (fine) and D (duplex) in one and of M (medium) and C (coarse) in the other. The F and D microstructures have higher yield strength in the entire test temperature range for both strain rates (Fig. 2). Since they both have fine size precipitates and F shows slightly higher yield strength than D, this could be related partially to the precipitate size effect. A D microstructure, having similar medium size cuboidal precipitates as M and fine size precipitates as F, might be expected to show somewhat an intermediate behavior. However, the existence of fine precipitates as in F seems to force D to behave more like F and the strength is lowered only slightly by the presence of the medium-size precipitates. This clearly illustrates the significant effect of the fine precipitates in the microstructure on the increase of the flow stress level. One of the strengthening mechanisms in precipitated microstructures is the blockage of moving dislocations by the precipitate particles. Therefore, it is expected to have better strengthening when there are finer precipitates and the interparticle spacing is shorter; it is interesting to note that this behavior persists even when the fine ones are found mixed with somewhat coarser particles, as in the duplex size microstructure.

In this study, an anomalous behavior of flow stress of the superalloy IN738LC is also observed at both the strain rates, and the results indicate an important effect of strain rate on the yielding characteristic. Comparison of Fig. 2(a) and (b) clearly shows that the anomalous increase in the yield strength has been pushed to a higher temperature at the higher strain rate for all the microstructures except F. It is also seen that the magnitude of the peak strength at 850 °C is higher (about 30%) with the higher strain rate than the analogous strengths of these microstructures at 850 °C with the lower strain rate. This transition apparently indicates that cross-slip occurs more easily at higher temperatures with higher strain rates, and it also corresponds to a change in deformation mechanism from dislocation cutting the γ' phase to formation of stacking faults coupled with antiphase domains,[8] which exert further strengthening, whereas a climb-forced bypassing mechanism leading to a decrease of strength prevails at the lower strain rates in the range 750 to 850 °C. A similar result was reported in Ref 5, 9, and 10.

The increase in yield strength has been observed at high temperatures at both strain rates in the γ'-strengthened superalloys earlier.[3] As stated before, the first observation of this anomalous rise in yield strength was reported by Westbrook[1] over 40 years ago. Also, the effects of γ' volume fraction and γ' size effect on the flow stress have been discussed in the literature in detail. In addition, the results reported here indicate the controlling effect of precipitate morphology on yielding. Investigation on the misfit of precipitates with fee matrix and the preferred orientations in the microstructures (Table 2) indicates that both the matrix and the precipitate phases have nearly identical lattice parameters and the misfit is nearly zero.[11] This implies that misfit-related stresses are not the controlling factor for the strength increase. Results in Fig. 2(a) and (b) show that the levels of yield strength are controlled by preferred orientation rather than precipitate sizes in the materials with the precipitate sizes of 700 and 450 nm. However, the same figures indicate that the effect of sizes dominates when the precipitate size is reduced to 70 nm in the F material and to 50 nm in the D material. It is concluded that the fine precipitate size and preferred orientations play a dominant role on the level of yield strength, even in the duplex precipitate material D with a mixture of precipitates of sizes 50 and 450 nm.

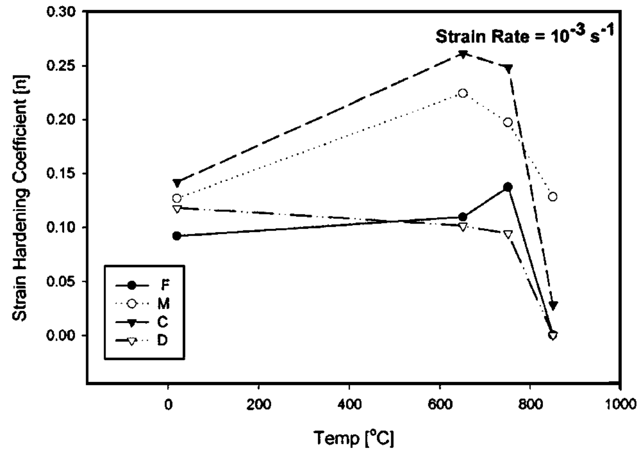


(a)

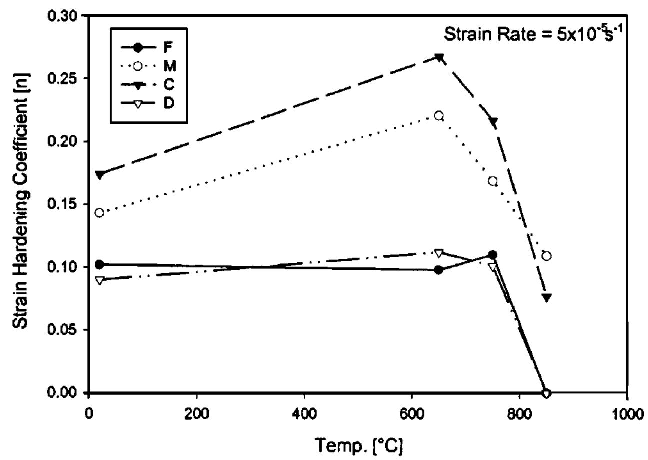


(b)

Fig. 3 Variation of ultimate tensile strength vs temperature at (**a**) high strain rate (10^{-3} s^{-1}) and (**b**) low strain rate (5×10^{-5} s^{-1})



(a)



(b)

Fig. 4 Variation of strain hardening (*n*) vs temperature at (**a**) high strain rate (10^{-3} s^{-1}) and (**b**) low strain rate (5×10^{-5} s^{-1})

4.2 Tensile Strength and Strain Hardening

The tensile strength behavior of the microstructures is shown in Fig. 3(a) and (b). There are again two distinct patterns based on precipitate sizes and combinations, one for F and D and the other for M and C precipitate microstructures. The F and D materials have slightly higher tensile strength at room temperature than the C and M in both strain rates. These materials also show a continuous decrease in strength when heated. However, it is obvious that the microstructures C and M pick up strength in the range Room Temperature to 650 °C and become slightly stronger than the F and D. Although, the increase or drop in tensile strength in the range Room Temperature to 750 °C is only about 10%, the drop is considerable for the range 750 to 850 °C. This drop is about 30% for all the microstructures under low strain rate, as opposed to the drops at the higher strain rate of about 10 to 15% only. These observations need to be explained further based on dislocation interactions and strain hardening tendencies. In fact, C, being the most strain-hardened microstructure, shows the highest tensile strength at 650 °C. Microstructure F behaves differently compared to the other microstructures. It not only strain hardens the least, but also shows an anomalous increase in strain hardening in the range 650 to 750 °C. The reason for this behavior is also not obvious, but in some way, it could be related to anomalous strengthening observed through the dislocation substructure in this temperature range. The aforementioned behavior indicates that the major role on the strengthening beyond the yield point is played by precipitate sizes and morphology rather than initial preferred orientation. It is seen that both C and M materials, having cuboidal and semicuboidal shapes, respectively, harden through plastic deformation and sustain their strength better in the high-temperature range studied. Although this is the case, the duplex size precipitate microstructure seems to prevail with the highest strength at the highest temperature (850 °C) investigated.

5. Conclusions

Based on the results obtained in this study, the following conclusions can be drawn.

- The tensile properties of IN738LC with different precipitate microstructures arrange themselves in two general groups—those of microstructures with the coarse (C) and medium (M) size precipitates and those of microstructures with the fine (F) and duplex size (D) precipitates. The behavior of microstructures in a group is nearly similar.
- The yield strength shows an anomalous increase in the range 650 to 750 °C at the lower strain rate, but with the higher strain rate, the anomaly occurs in the range 750 to 850 °C. However, it appears that the yield strength is effected primarily by the precipitate size in D and F precipitate materials.
- The tensile strength of C and M microstructures seems to increase slightly in the range Room Temperature to 650 °C, which seems to be related to an increase in strain hardening in this temperature range for these microstructures. The results indicate that both morphology and size of precipitates effect the resultant tensile strength.
- The reasons for the varied behavior of the different microstructures should be associated with the dislocation - precipitate interactions at the different temperatures. The anomalous strengthening known to occur in Ni_3(Al, Ti)-type intermetallic compounds could be partially responsible for the anomalous strengthening in IN738LC at 750 C or 850 °C for different microstructures.

Acknowledgments

This study was supported financially by NASA through the LaSpace program in the State of Louisiana. Howmet Corp. extended cost consideration for specimens, for which the authors are grateful. EB thanks Mugla University (Mugla, Turkey) for financial support toward his doctoral studies at LSU.

References

1. J.H. Westbrook: *Trans. AIME,* 1957, vol. 209, p. 898.
2. N.S. Stoloff and C.T. Liu: in *Physical Metallurgy and Processing of Intermetallic Compunds,* N. S. Stoloff and V. K. Sikka, eds., Chapman & Hall, New York, NY, 1994, pp. 159-296.
3. D.L. Anton: in *Intermetallic Compounds: Principles and Practice,* J.H. Westbrook and R.L. Fleischer, eds., John Wiley & Sons, New York, NY, 1995,] pp. 3-15.
4. A.J. Ardell: in *Intermetallic Compounds: Principles and Practice,* ibid, pp. 257-86.
5. D. Bettge, W. Osterle, and J. Ziebs: *Z. Metallkd.,* 1995, vol. 86 (3), pp. 190-97.
6. E. Balikci, A. Raman, and R.A. Mirshams: *Metall. Mater. Trans. A,* 1997, vol. 28A, pp. 1993-2003.
7. E. Balikci, R. A. Mirshams, and A. Raman: *Mater. Sci. Eng. A,* 1999, vol. A265, pp 52-62.
8. D. Mukherji, F. Jiao, W. Chen, and R.P. Wahi: *Acta Metall. Mater.,* 1991, vol. 39 (7), pp. 1515-24.
9. D.P. Pope and S. S. Ezz: *Int. Met. Rev.,* 1984, vol. 29 (3), pp. 136-67.
10. M.F. Kniepmeier, T. Link, I. Poschmann, G.S. Freker, and C. Schulze: *Acta Metall. Mater.,* 1996, vol. 44 (6), pp. 2397-2407.
11. E. Balikci, R.E. Ferrell, Jr., and A. Raman: *Z. Metallks,* 1999, vol. 90 (2), pp. 141-46.
12. D.A. Porter and K.E. Easterling: *Phase Transformation in Metals and Alloys,* 2nd ed., Chapman & Hall, New York, NY, 1993, p. 316.