TWIN BOUNDARY CAVITATION DURING HIGH TEMPERATURE CREEP OF INCONEL ALLOY X-750

M. C. Pandey* and D. M. R. Taplin**

**Department of Mechanical Engineering, University of California, Davis, CA 95616, USA*
***Department of Mechanical and Manufacturing Engineering Trinity College, University of Dublin, Dublin-2, Ireland*

ABSTRACT

Evidence of cavities/cracks along twin boundary has been found in Inconel alloy X-750 tested under constant load creep and constant cross-head speed tensile conditions at elevated temperatures. These twin boundary cavities/cracks are mostly observed near the fracture surface and in those specimens which failed after relatively large elongations.

KEYWORDS

Twin boundary cavitation; nickel-base superalloy; high temperature; creep ductility; creep fracture.

INTRODUCTION

When polycrystalline engineering materials are subjected to creep in the temperature range of 0.4 Tm to 0.7 Tm (where Tm is the melting point in absolute scale), cavities often nucleate along the grain boundaries. Growth and interlinkage of these cavities cause intergranular embrittlement and fracture. However, under certain material and testing conditions, cavities can also nucleate (grow) along twin boundaries. For example, Sikka, Swindeman and Brinkman (1977) observed a transition from grain to twin-boundary cavitation in type 304 and 316 stainless steel specimens when aged and creep tested at 593°C. Evidence of twin boundary cavitation was also observed in unaged specimens of stainless steel tested at 649°C at certain stresses. Mitchel, Nahm and Moteff (1973) observed cavities at both grain and twin boundaries in type 304 stainless steel tested at 650°C. In the case of nickel-base superalloys, cracks along twin boundaries have only been reported in Inconel 718 tested under low cycle fatigue condition (Fournier and Pineau, 1977). The purpose of this paper is to report on twin boundary cavitation in Inconel alloy X-750 tested under constant load creep and constant cross-head speed tensile conditions at elevated temperature.

EXPERIMENTAL PROCEDURE

The material used in this investigation is Inconel alloy X-750; the nominal composition is as follows (wt.): 0.05 C, 14.9 Cr, 6.98 Fe, 2.5 Ti, 0.61 Al, 0.92 (Nb + Ta), 0.1 Mn, 0.18 Si, 0.01 S and balance Ni. The material was given the standard industrial heat-treatment: 1150°C for 4h, air cool; 840°C for 24h, air cool and 710°C for 20h, air cool. This resulted in an average grain size of 160 µm, measured by the linear intercept method. Constant load creep and constant cross-head speed tensile tests were carried out at 700°C on specimens of diameter 5mm with a gauge length of 25mm. A few tensile tests were also carried out at 550°C, 625 and 800°C. The specimens were mid sectioned, polished and electroetched in a 5 percent nital solution for microstructural examinations. Optical, scanning and field emission scanning electron microscopes were used to study the cavities and fractured surfaces.

RESULTS

Constant load uniaxial creep test was carried out at 390 MPa and at a temperature of 700°C on a specimen which was heat-treated in bar form of size 12.5mm x 12.5mm in air. The heat-treatment in the bar form was done to avoid any environmental interactions prior to testing (Pandey, 1982). The specimen fractured after 283 hours at percentage elongation and reduction in area of 9 and 20, respectively. Figure 1 shows the crack along the twin boundary near the fracture end of the specimen. The twin boundary crack is at approximately 45° to the stress axis. Further, it may be observed that there is some amount of the grain matrix (intragranular) deformation.

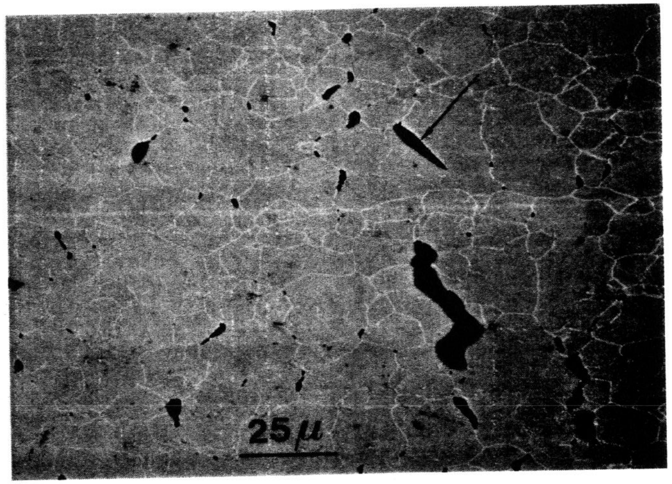

Fig. 1. Twin boundary crack (tbc) observed near the fracture end of the specimen creep tested at the stress of 390 MPa and at 700°C. Note the elongated grains in the near region of the twin boundary crack (stress axis horizontal).

Another example of twin boundary cavitation is shown in Figure 2a. This looks like twin boundary at the lower magnification. But at the higher magnifications, it gives convincing evidence of being cavities (Fig. 2b). It is further noted that there are no precipitate free zones along the twin boundaries (they are present along grain boundaries). Further example of the twin boundary cracks is shown in Figure 3. This is obtained from a specimen which was tested at 400 MPa and at 700°C. The specimen failed after 360 hours at percentage elongation of 12 and percentage reduction in area of 24. In this case also, it may be noted that the twin boundary cracks are at 45° to the stress axis and close to the fracture end. No twin boundary cracks/cavities were observed in those specimens which were heat-treated at a reduced pressure of 2.60×10^{-2} Pa and fractured giving very poor creep ductility (percentage elongation and reduction in area of 1 and 7, respectively) (Pandey, Dyson and Taplin, 1984).

The twin boundary cracks were also observed in the tensile tested specimens at a cross-head speed of $0.054\text{mmh}^{-1}(2.16\times10^{-3}\text{h}^{-1})$ at 700°C (Fig. 4). Two specimens were tested: one was interrupted prior to its final fracture (percentage elongation and reduction of area 11.90 and area 21, respectively); the other one was allowed to fracture at percentage elongation and reduction in area of 13 and 24, respectively. Some of the cracks are lying on those twin boundaries which are bent and are at 45° to the stress-axis. As compared to the interrupted one, not many twin boundary cavities/cracks were observed in the fractured specimen. This appears to indicate that the twin boundary cracks are part of the fracture path (Fig. 5a and b). Scanning electron microscopy of the fracture surfaces indicated two modes of fracture: (i) intergranular and (ii) faceted as reported in other nickel-base superalloys (Mills, 1980; Oblak and Owczarski, 1972). It therefore appears that the faceted zones are in fact twin boundaries. Tensile testings carried out at 550°C, 625°C and 800°C at the cross-head of 0.054mmh^{-1} showed the twin boundary cracks only in the specimen tested at 800°C. A plot between creep ductility and temperature shows that the ductility minimum occured at 625°C (Pandey, Taplin and Mukherjee, 1984) (Fig. 6). It was found that very low ductility obtained at 550°C and 625°C are due to oxidation at grain boundaries. The specimen tested at 800°C was interrupted prior to its final fracture. Twin boundary cracks were observed just ahead of the growing crack tip.

DISCUSSION

The results presented above show that cavities can also nucleate and grow along the twin boundaires under certain testing conditions. An important point to note is that the majority of the twin boundary cavities/cracks lie at approximately 45° to the stress-axis. Therefore, it seems that the cavities are nucleated due to sliding since maximum shear deformation occurs on those twin boundaries which lie at 45° to the stress axis. Dunlop and Howell (1983) suggest that during dislocation creep, extrinsic grain boundary dislocations are formed and their movement causes sliding at the twin interface. Wirmark, Nilsson and Dunlop (1981), in austenitic stainless steel, observed that the amounts of sliding at the twin interface were proportional to the total creep strain. Twin boundary cavities/cracks observed near the fracture end of the alloy Inconel X-750 appear to indicate that sliding along the

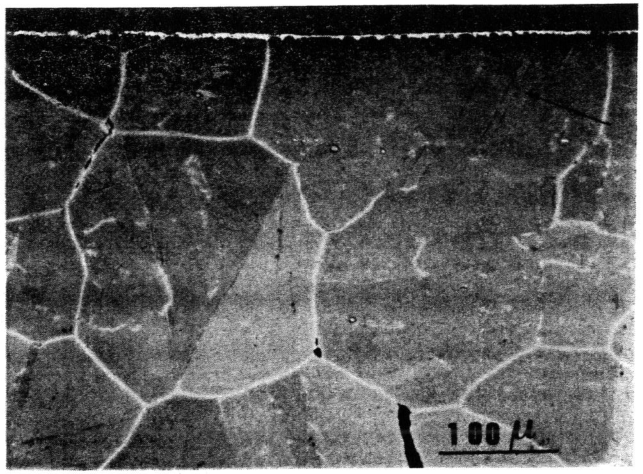

Fig. 2a. Note the twin boundary cracks (shown by arrow) near the specimen's surface and close to the fracture end. The specimen is the same as in Figure 1 (stress axis horizontal).

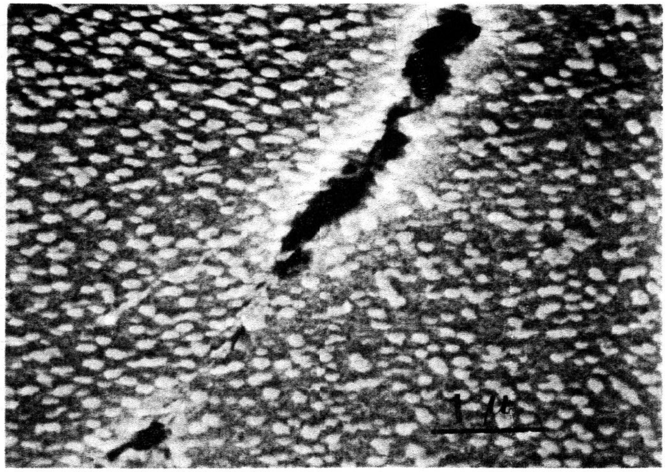

Fig. 2b. Enlargement of the twin boundary cavities of the photomicrograph shown in Fig. 2a. by Field Emission Scanning Electron Microscopy.

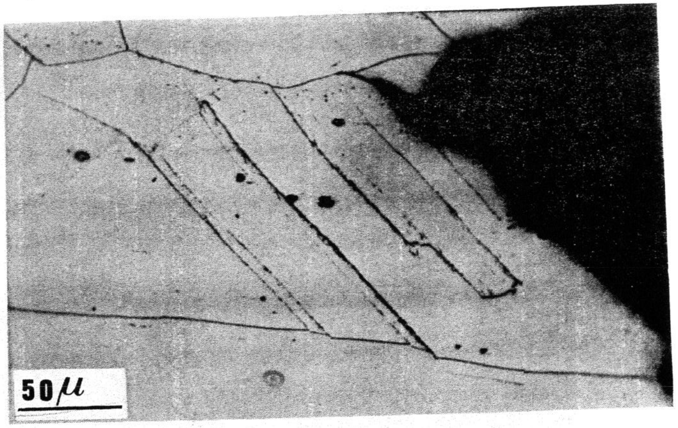

Fig. 3. Twin boundary cracks observed near the fracture surface of the specimen creep tested at 400 MPa and at 700°C. The specimen fractured after 360 hours giving percentage elongation and reduction of area 12 and 24 respectively (stress axis horizontal).

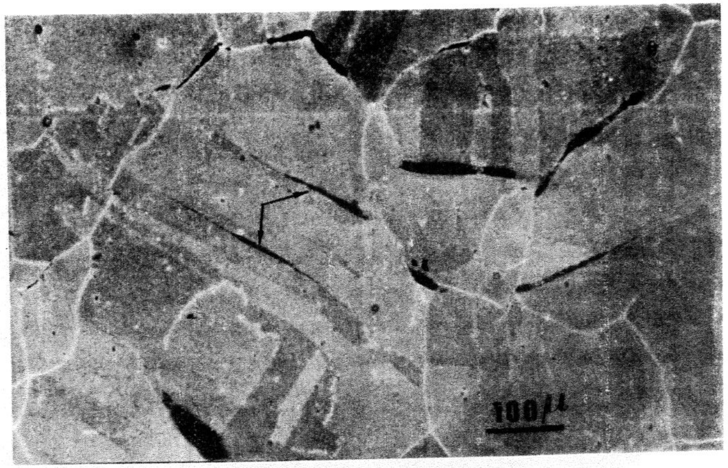

Fig. 4. Twin boundary cracks in the necked region of an interrupted specimen, heat-treated in air and tensile tested at 700°C at a constant cross-head speed of $0.054 \mathrm{mmh}^{-1}$ (stress axis vertical).

Fig. 5a. Showing twin boundary cracks in the specimen which was heat treated and tested under identical conditions to that of Fig. 4 but interrupted prior to fracture. The specimen would have fractured along the twin boundary if allowed to fail (stress axis vertical).

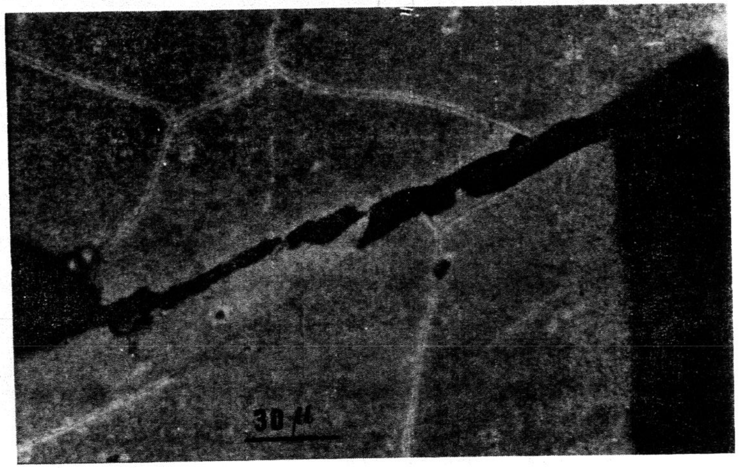

Fig. 5b. Enlargment of the segment of the twin boundary crack of Figure 5a.

twin interface caused cavities to nucleate. Therefore, two factors seem to play important roles for crackings to occur along the twin boundaries: (i) sliding along the twin boundaries and (ii) extensive intragranular deformation. For example, when the material was embrittled either by prior environmental interactions during heat-treatment (Pandey, Dyson and Taplin, 1984) or by prior room temperature prestraining (Pandey, Mukherjee and Taplin, 1984), no twin boundary cavities were observed in the creep tested specimens. In these specimens, intragranular deformation was very much restricted and failure occured after one percent of creep strain. Similarly, no twin boundary cracks were observed in those specimens which were tested at 550°C and 625°C at the constant cross-head speed. The specimens failed showing poor ductility (Fig. 6); slidings along the twin and grain boundaries were very much restricted due to the temperatures being low.

Extensive intragranular deformation appears to cause more twinning, as the pressure of large numbers of twin boundaries near the fracture end indicated. Chin, Hosford and Mendorf (1969) have observed that twinning always occured on the most stressed region of the tested specimen. Twinning also causes an increase in the flow stress since a high stress concentration occurs at the twin boundaries which are not completely relieved and numerous dislocation debris exist (Remy, 1981). Mills (1980) investigated deformation and fracture characteristics of Inconel alloy X-750 at temperatures ranging from 24°C to 816°C at constant strain rate. The fracture surfaces exhibited extensive faceting in addition to intergranular cracking in the temperature range of 538°C to 704°C. Mills believed that faceted fracture surface was associated with dislocation channels formed by an extensive hetrogeneous slip mechanism and ignored twin boundaries. However, it now seems quite clear that such faceted fracture surfaces are associated with twin boundaries.

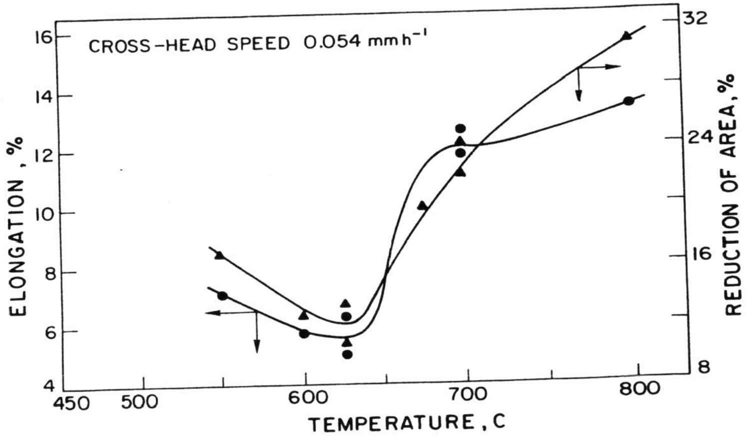

Fig. 6. Creep ductility vs testing temperature in specimens, heat-treated in bar form and tensile tested at a cross-head speed of 0.054mmh^{-1}. Note ductility minimum at 625°C.

CONCLUSION

The following conclusions can be drawn from the present investigation on Inconel alloy X-750:

1. Cavities can nucleate (grow) along twin boundaries at elevated temperatures probably by a twin boundary sliding mechanism.
2. The twin boundary cracks as well as the elongated grains observed close to the fractured surface appears to indicate that local matrix (intragranular) deformation is a necessity for cavities to nucleate along the twin boundaries.
3. Cracks along the twin boundaries occurred in those specimens which fractured after relatively large elongations.
4. The faceted fracture surfaces appear to be associated with the twin boundaries

ACKNOWLEDGEMENT

The authors would like to express their gratitude to the EXXON Corporation and the Department of Energy, Mines and Resources, Canada for the financial support. Discussions with Dr. G. L. Dunlop and Dr. B. F. Dyson are acknowledged with pleasure.

REFERENCES

Chin, G. Y., W. F. Hosford and D. R. Mendorf (1969). Proc. Roy. Soc. A, London, 309, 433-456.

Dunlop, G. L. and P. R. Howell (1983). In M.H. Lewis and D.M.R. Taplin (Ed.), Micromechanisms of Plasticity and Fracture, University of Waterloo Press and Parson Press, Trinity College, Dublin, 77-106.

Fournier, D. and A. Pineau (1977). Metall. Trans. A, 8A, 1095-1105.

Mills, W. J. (1980). Metall. Trans. A, 11A, 1039-1047.

Mitchel, D. J., H. Nahm and J. Moteff (1973). Mat. Sci. Eng., 11, 97-102.

Oblak, J. A. and W. J. Owczarski (1972). Metall. Trans., 3, 617-626.

Pandey, M. C. (1982). Ph.D. Thesis, University of Waterloo, Canada.

Pandey, M. C., B. F. Dyson and D.M.R. Taplin (1984). Proc. Roy. Soc. A, London, to be published.

Pandey, M. C., D.M.R. Taplin and A. K. Mukherjee (1984). Metall. Trans. A to be published.

Pandey, M. C., A. K. Mukherjee and D.M.R. Taplin (1984). Metall. Trans. A, to be published.

Remy, L. (1981). Metall. Trans. A, 12A, 387-408.

Sikka, V. K., R. W. Swindeman and C. R. Brinkman (1977). Metall. Trans. A, 8A, 1117-1129.

Wirmark, G., J.-O. Nilsson and G. L. Dunlop (1981). Phil. Mag. A, 1, No. 43, 93-101.