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THE FRACTURE SURFACE MICROMORPHOLOGY OF INCONEL X-750 AT ROOM TEMPERATURE AND ELEVATED TEMPERATURES

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ABSTRACT

The fracture surface micromorphology of Inconel X-750 has been examined in order to relate operative fracture mechanisms to key microstructural features. Under room temperature and intermediate temperature (approximately 600°F) conditions, failure occurred primarily by an intergranular dimple rupture mechanism associated with microvoid coalescence along a grain boundary denuded region. At progressively higher temperatures an intergranular to transgranular fracture mechanism transition was detected. In the 1000 to 1300°F regime, the Inconel X-750 fracture surfaces were dominated by a faceted transgranular morphology. This faceted appearance was attributed to extensive heterogeneous planar slip, characteristic of many nickel-base superalloys.

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THE FRACTURE SURFACE MICROMORPHOLOGY OF INCONEL X-750 AT ROOM TEMPERATURE AND ELEVATED TEMPERATURES

INTRODUCTION

Inconel* X-750 (AISI 688), like many nickel-base superalloys, has a low stacking fault energy fcc matrix (γ) and is strengthened through precipitation of $Ni_3[Al,Ti]$, termed gamma prime (γ'), and various metallic carbides. This nickel-base superalloy possesses excellent corrosion and oxidation resistance in addition to high strength and ductility at temperatures up to 1300°F. As a result of these properties, Inconel X-750 has been considered for application as a heat-resistant structural material in both the nuclear reactor and aerospace industries.

The continued safe operation of Inconel X-750 structural components requires a comprehensive understanding of the overall fracture behavior and basic fracture mechanisms. Consequently, the current investigation was undertaken to characterize the room temperature and elevated temperature fracture surface appearance of Inconel X-750 and to relate operative fracture mechanisms to key microstructural features found in this nickel-base superalloy. The work discussed herein summarizes metallographic and fractographic observations on Inconel X-750 tensile specimens tested at temperatures ranging from 75 to 1300°F.

EXPERIMENTAL PROCEDURE

Material and Mechanical Testing

The Inconel X-750 used in this study was obtained in the form of a 1-inch thick plate. This material was from heat HT6716X produced by the Huntington Alloy Products Division of the International Nickel Company using an air-melt process. The chemical composition of this heat of Inconel X-750 is given in Table I, and a detailed thermal-mechanical processing history of this plate is given in Reference 1.

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* Inconel is a registered trademark of the International Nickel Company. General specifications for this material may be found in ASTM A-637.

TABLE I

CHEMICAL COMPOSITION (percent by weight)

Producer/Heat No.	C	Mn	Fe	S	Si	Cu	Ni	Cr	Al	Ti	Co
Huntington/HT6716X	0.03	0.56	6.66	0.007	0.40	0.09	62.84	15.27	0.67	2.60	0.07

Melt process: air-melt

Heat treatment: Solution anneal 2100°F for 2 hours and air cool. Age 24 hours at 1550°F and air cool. Age 20 hours at 1300°F and air cool.

Miniature buttonhead tensile specimens (Figure 1) were fabricated from the longitudinal forming direction of the plate and subsequently tested in the precipitation hardened condition. The double aging heat treatment for Inconel X-750 is given below:

- solution annealed at 2100°F for 2 hours, air cooled
- aged at 1550°F for 24 hours, air cooled
- aged at 1300°F for 20 hours, air cooled.

Tensile tests were performed on an Instron Tensile Machine (nominal strain rate of 1.78×10^{-3} min^{-1}) at temperatures ranging from room temperature to 1300°F in accordance with ASTM Specification E21-70[2].

Electron Microscopy Techniques

Surface replicas were prepared from solution annealed and double aged samples in order to examine the precipitate morphology of Inconel X-750. The surface of each sample was mechanically polished and electrolytically etched in an oxalic acid solution. Standard two-stage (Pt-Pd-C) replicas were then prepared from each polished surface and examined on a Phillips electron microscope operated at an accelerating potential of 60 kV.

The fracture surface appearance of this nickel-base alloy was characterized by fractographic examination of tensile specimens on an AMR Model 1200 scanning electron microscope at an accelerating potential of 15 kV.

RESULTS AND DISCUSSION

Microstructure

The typical microstructure of the precipitation hardened Inconel X-750 is illustrated in Figures 2a and 2b. Double aging heat treatment resulted in the formation of a duplex γ' precipitate structure in the grain interior as coarse γ' particles appeared after the 1550°F intermediate age, and fine γ' particles formed during the final 1300°F age. In addition to the γ' precipitate structure, Figure 2a reveals the presence of a chromium-rich $M_{23}C_6$ carbide precipitate[3] along the grain boundary accompanied by a zone on either side of the boundary depleted in coarse γ' particles. Higher magnification

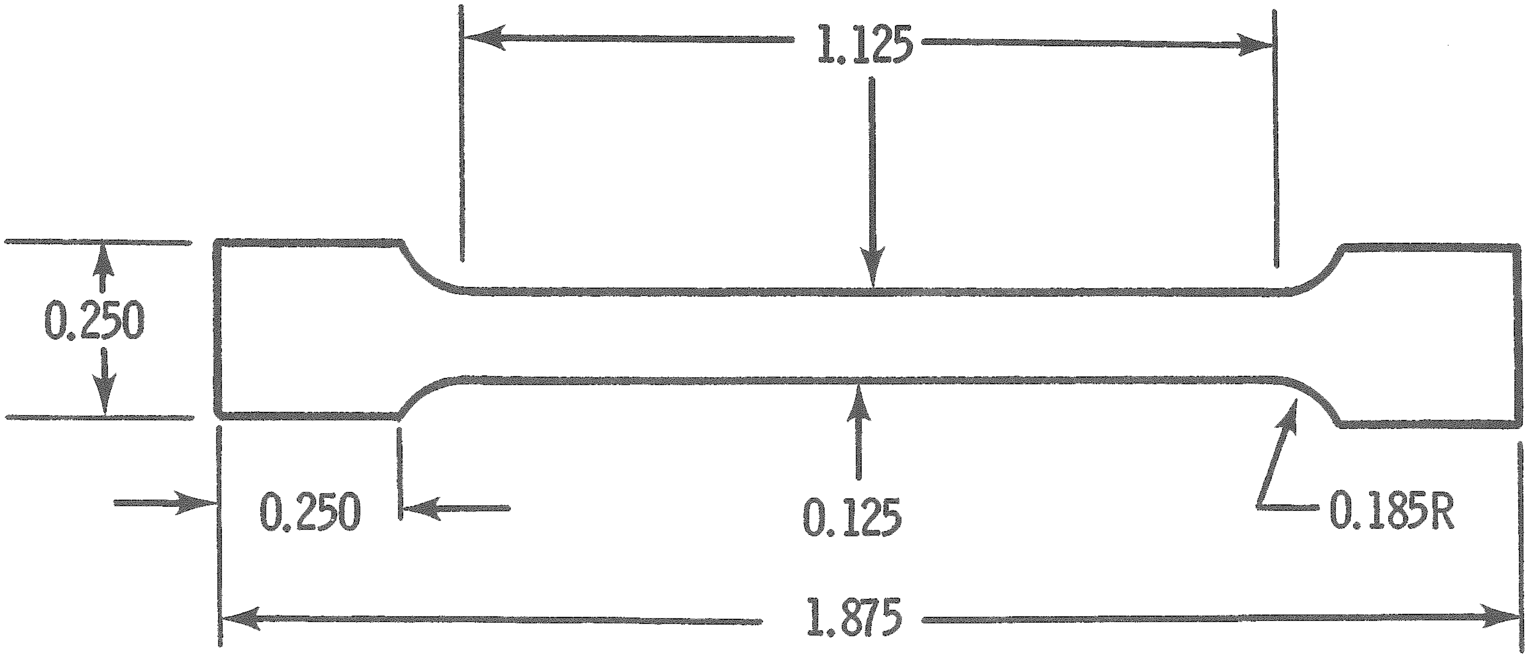


Figure 1 - Tensile Specimen Geometry (Dimensions in inches).

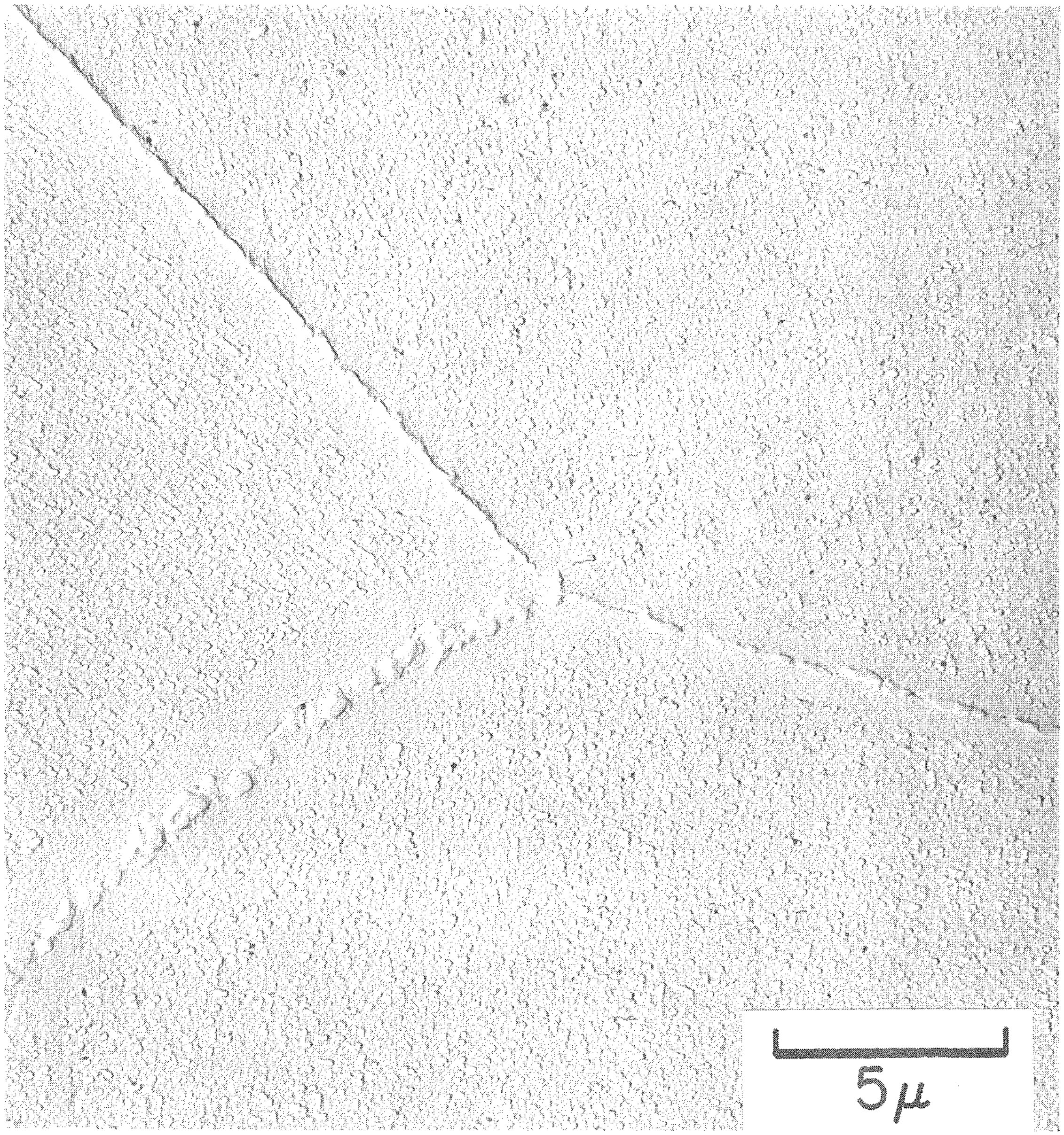


Figure 2a - Microstructure of the precipitation heat treated Inconel X-750 illustrating the duplex γ' precipitate morphology and the chromium-rich $M_{23}C_6$ carbides decorating the grain boundaries.

of the grain boundary region (Figure 2b) illustrates, however, that fine γ' particles had precipitated within this denuded zone, thereby eliminating the formation of a totally depleted region. The net result is the presence of narrow zones along grain boundaries that are somewhat weaker, but in general more ductile, than the grain interior.

Depletion of the coarse γ' is believed to be associated with chromium diffusion to the grain boundary creating zones impoverished in chromium[4,5]. This decreased chromium content allows complete solubility of titanium and aluminum within the γ matrix at 1550°F; hence, no coarse γ particles form along the chromium-depleted grain boundary regions during the intermediate 1550°F age. At the lower aging temperature, 1300°F, the solubility of titanium and aluminum is sufficiently reduced so that fine γ' particles can precipitate within this denuded zone.

Tensile Behavior

Room temperature and elevated temperature tensile property data for precipitation heat treated Inconel X-750 are tabulated in Table II, and the effect of temperature on the strength and ductility of this nickel-base superalloy is illustrated in Figure 3. The 0.2% offset yield strength was found to be relatively insensitive to temperature decreasing from 92.4 ksi at room temperature to 80.3 ksi at 1300°F. On the other hand, a rather significant reduction in ultimate strength was observed at temperatures above 1000°F. Both the elongation and reduction in area were found to reach a maximum value at approximately 800°F. These tensile property trends for precipitation hardened Inconel X-750 are consistent with those reported in Reference 6.

The fracture surface morphology of the Inconel X-750 tensile specimens was examined in order to relate fracture mechanisms with key microstructural features. Under room temperature conditions, failure occurred primarily by an intergranular fracture mechanism as shown in Figure 4. The rumpled appearance found on the intergranular facets was attributed to a dimple rupture network superimposed on the grain walls. Similar intergranular dimple rupture mechanisms, generally associated with coalescence of voids around grain boundary precipitate particles, have been reported in the literature[7-10]. In the

TABLE II

TENSILE PROPERTIES*

Specimen Number	Test Temp.	0.2% Yield Strength	Ultimate Strength	Uniform Elongation	Total Elongation	Red. in Area	Site Sequence Number**
T106	75°F	93,090 psi	149,190 psi	14.2%	14.3%	15.4%	H05986
T125	75°F	91,800 psi	152,500 psi	18.2%	18.8%	17.2%	H05988
T107	600°F	85,660 psi	140,000 psi	13.9%	14.3%	23.0%	H05983
T527	600°F	89,240 psi	151,260 psi	20.7%	20.7%	26.0%	H05991
T108	800°F	82,100 psi	148,390 psi	18.8%	19.6%	25.8%	H05984
T123	800°F	83,200 psi	142,400 psi	21.0%	21.2%	23.8%	H05990
T109	1000°F	81,560 psi	134,840 psi	15.9%	16.9%	21.3%	H05982
T124	1000°F	80,000 psi	130,900 psi	15.2%	15.7%	18.0%	H05989
T110	1200°F	79,670 psi	108,030 psi	5.3%	5.8%	13.1%	H05987
T528	1200°F	81,970 psi	106,970 psi	5.5%	5.6%	9.8%	H05992
T111	1300°F	78,460 psi	91,710 psi	3.5%	4.2%	11.4%	H05981
T523	1300°F	82,230 psi	93,720 psi	5.2%	5.4%	5.8%	H05993

* Strain rate = 1.78×10^{-3} min^{-1}

** HEDL LMFBR Fuel Cladding Information Center

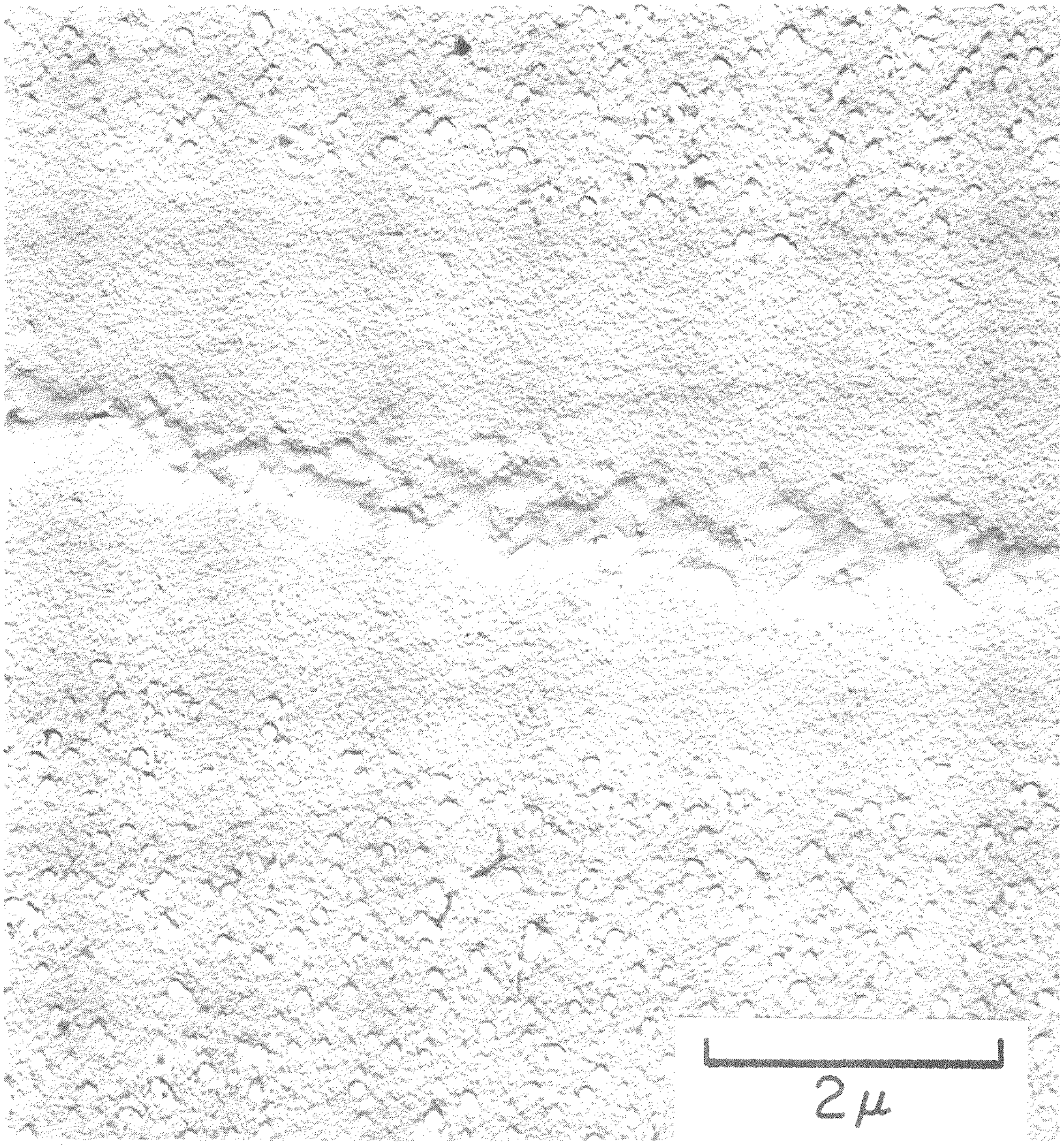


Figure 2b - Microstructure of the precipitation heat treated Inconel X-750 revealing the presence of a zone on both sides of the grain boundary that is depleted in coarse γ' precipitates. Note that the smaller γ' particles have precipitated within this denuded region.

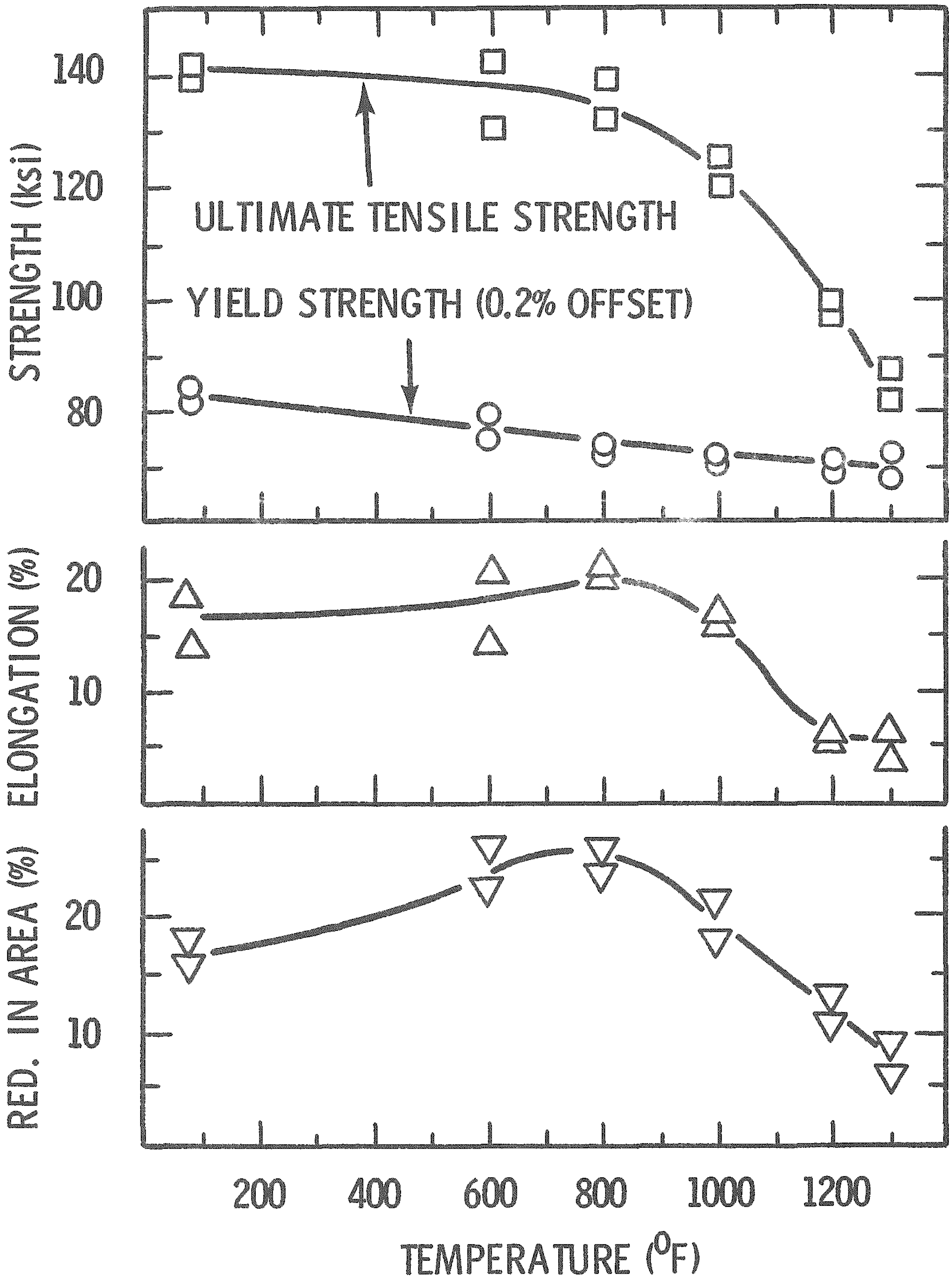


Figure 3 - Effect of temperature on the tensile properties of precipitation heat treated Inconel X-750.

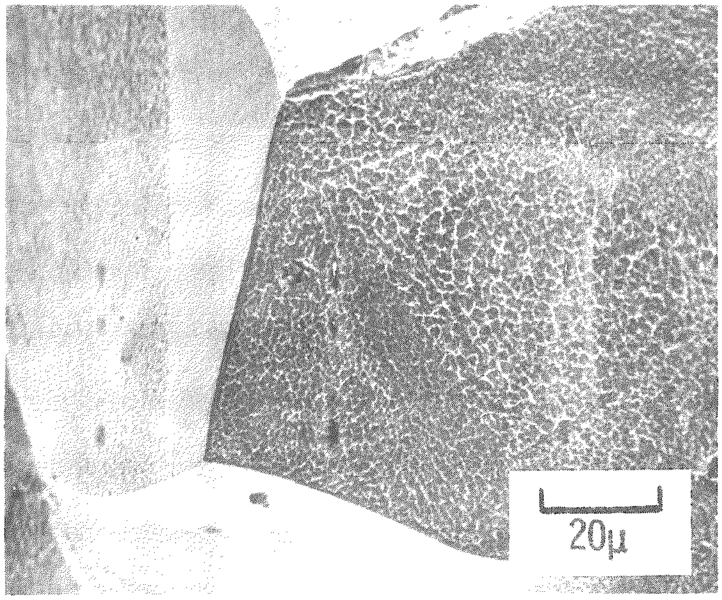
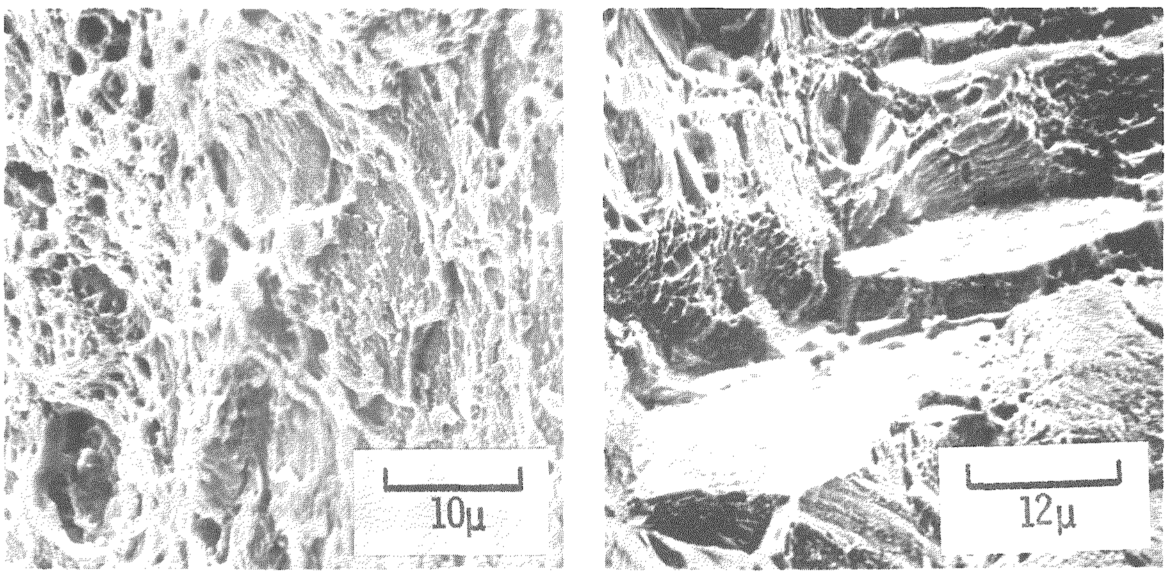


Figure 4 - SEM fractograph showing the intergranular dimple rupture network observed at room temperature.



(a) (b)

Figure 5 - Electron fractographs revealing the fracture surface appearance at 800°F.

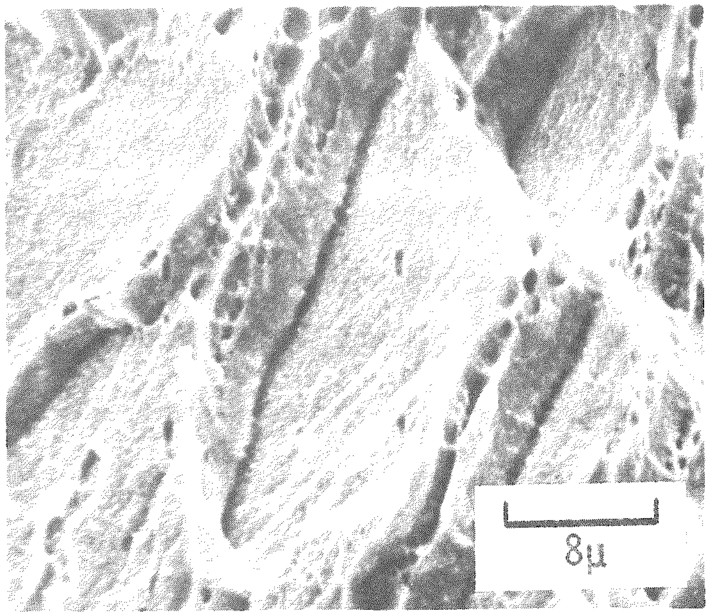
(a) Poorly defined dimples (on the left side of this fractograph) coupled with rather flat facets.

(b) Well defined planar facets.

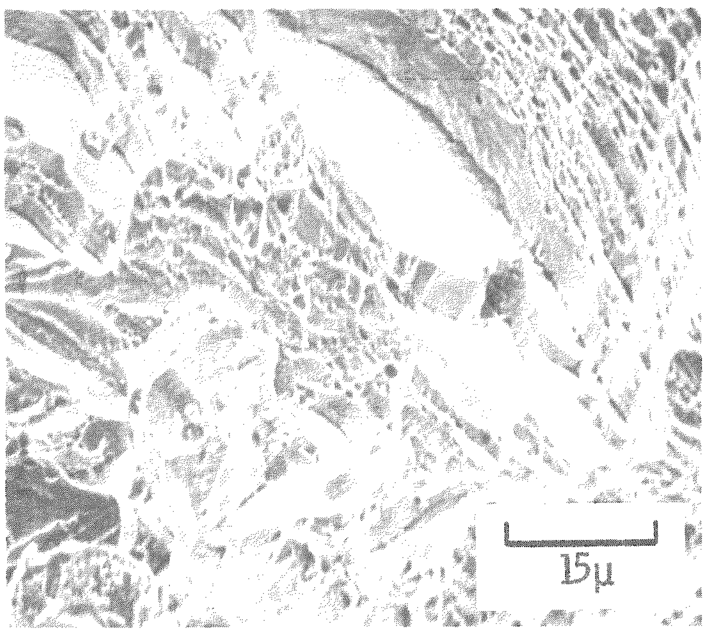
Inconel X-750 alloy, this intergranular fracture mode is believed to result from microvoid coalescence along the weakened grain boundary denuded zone (recall Figure 2b). Apparently, the rigid grain boundary carbides inhibit plastic relaxation of stresses which results in decohesion along the carbide-matrix interface. Subsequent growth and coalescence of these carbide-matrix microvoids within the grain boundary denuded region resulted in the observed intergranular dimple rupture network.

Under elevated temperature conditions, an intergranular to transgranular fracture mechanism transition occurred. At 600 and 800°F, approximately equal amounts of intergranular dimple rupture (shown in Figure 4) and transgranular (Figure 5) fracture modes were observed. The latter mechanism consists of poorly defined dimples coupled with a rather flat, faceted appearance as shown in Figure 5a. In addition, a limited number of planar ledges (Figure 5b) were detected on the 800°F fracture surface. Under higher temperature conditions, in the 1000 to 1200°F regime, the fracture surfaces exhibited a well defined faceted transgranular morphology as illustrated in Figures 6 and 7. Similar faceting mechanisms in nickel-base superalloys have been reported by Oblak et al.[11] and Menon et al.[12] Finally, at the highest test temperature, 1300°F, the operative fracture mechanisms were found to be a combination of intergranular dimple rupture (Figure 8a) coupled with transgranular faceting (Figure 8b). Note the well defined planar steps or ledges shown in Figure 8b.

These crisp, planar steps or facets illustrated in Figures 6a, 7 and 8b are reminiscent of a channel fracture mechanism [13,14] observed in neutron-irradiated Type 304 stainless steel. This mechanism is believed to result from a confinement of slip on a narrow band of planes such that the material literally slips apart[13]. In a similar fashion, the faceted fracture morphology observed in Inconel X-750 is also suspected to be associated with channels formed by extensive heterogeneous slip that is characteristic of nickel-base superalloys. More specifically, deformation of most γ' precipitation hardened alloys involves a planar slip mechanism where paired a/2<110> dislocations eventually shear both the γ matrix and γ' particles[11,15-19]. After a dislocation pair cuts through a region clearing a channel of sheared



(a)



(b)

Figure 6 - SEM fractographs of the 1000°F fracture surface illustrating the faceted transgranular morphology.

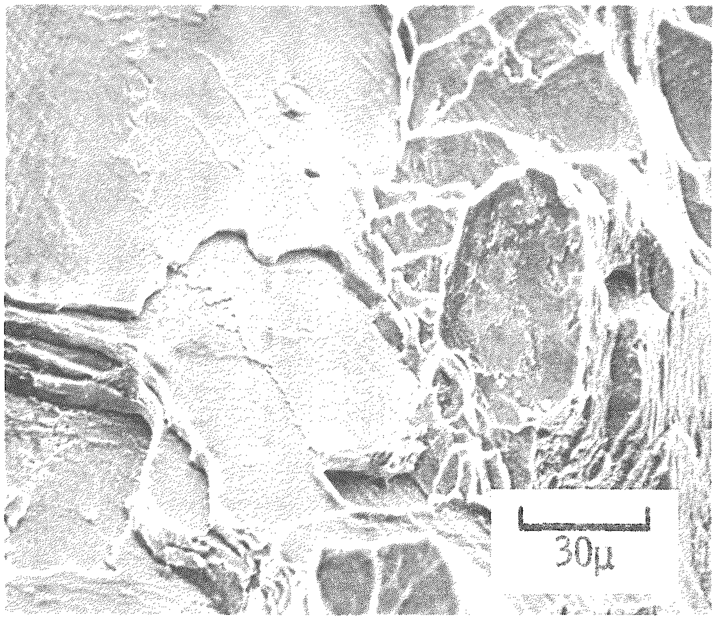
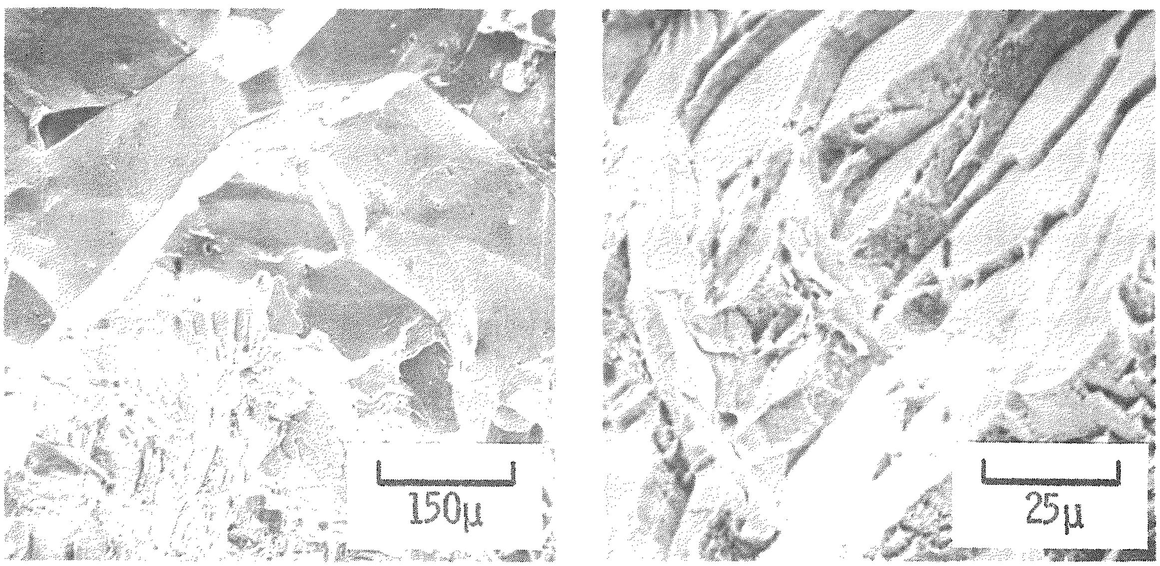


Figure 7 - Typical fractographic profile of the 1200°F fracture morphology.



(a) (b)

Figure 8 - Electron fractographs illustrating the fracture surface appearance at 1300°F.
(a) A combination of intergranular dimple rupture (upper portion of this fractograph) and transgranular faceting.
(b) Evidence of planar steps or facets within the transgranular region of Figure 8a.

γ' particles, the effectiveness of these sheared precipitates as dislocation barriers is diminished such that all subsequent slip is "channelled" through these zones. Consequently, the cohesive strength of the atomic bonds along these narrow channels is so weakened by extensive dislocation activity that the material literally slips apart (as illustrated in Figure 9), thereby causing the stepped or faceted fracture appearance termed channel fracture[13].

The fracture mechanism transition described above appears to have a significant effect on the practical engineering behavior of Inconel X-750. For example, it is noteworthy that the reduction in ultimate strength and ductility observed at temperatures above 1000°F (recall Figure 3) was accompanied by the onset of a well defined channel fracture mechanism caused by extensive dislocation activity on narrow slip bands. On the basis of these findings, it is reasonable to suspect that the degradation in tensile properties was associated with intense planar slip and the onset of channel fracture under elevated temperature conditions.

SUMMARY AND CONCLUSIONS

The results of the present fractographic study of Inconel X-750 tensile specimens tested at temperatures ranging from 75 to 1300°F may be summarized as follows:

- The intergranular dimple rupture network observed under room temperature and intermediate temperature conditions is believed to be associated with microvoid coalescence along the weakened grain boundary denuded zone.
- In the high temperature regime, an intergranular to transgranular fracture mechanism transition occurred. The transgranular channel fracture morphology observed at temperatures above 1000°F resulted from intense heterogeneous slip bands which initiated local separation.
- The degradation of tensile properties above 1000°F is believed to be associated with the onset of channel fracture.

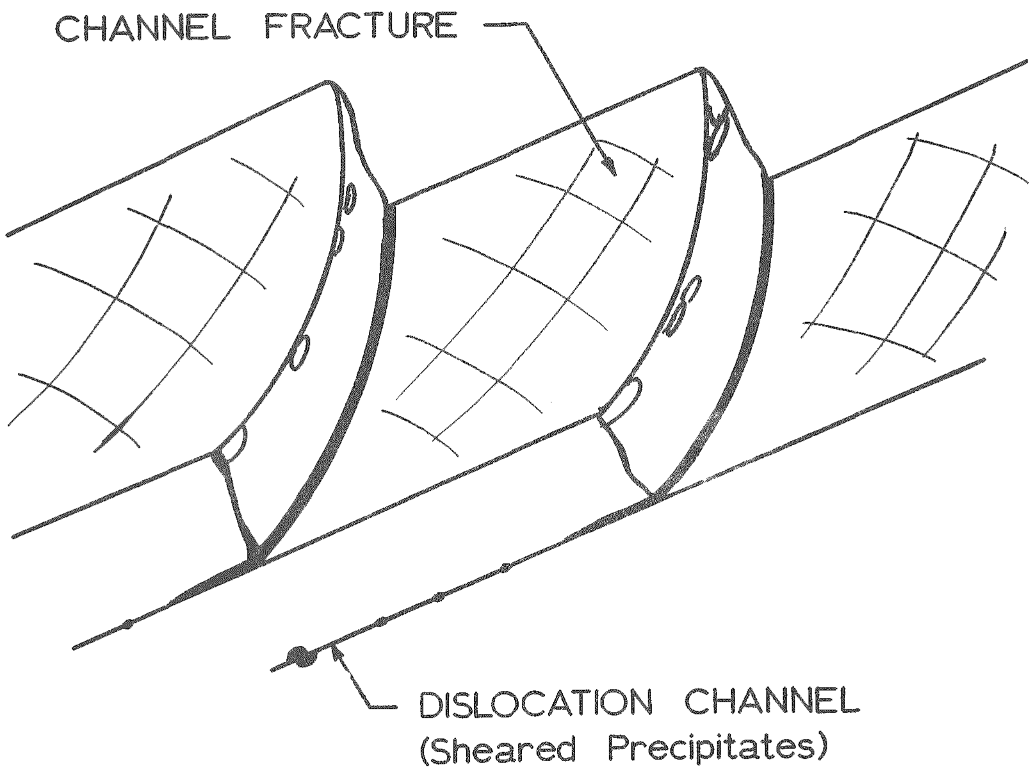


Figure 9 - Sketch of a model describing the channel fracture mechanism. Slip is concentrated along a narrow band of planes forming a dislocation channel. These zones are so weakened by extensive dislocation activity that material separation actually occurs along the dislocation channels; hence, the term channel fracture was chosen.

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