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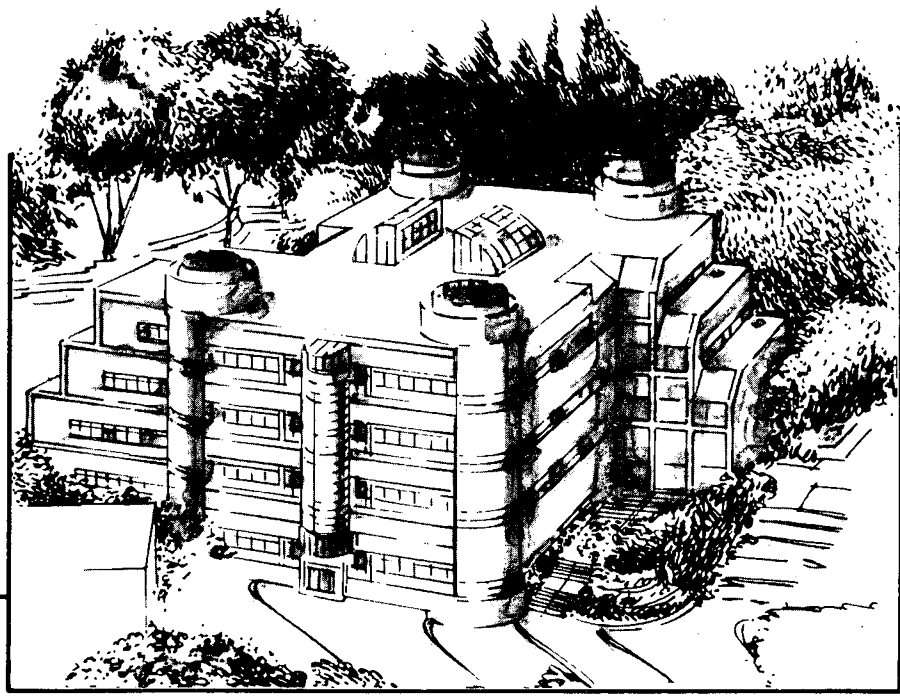

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INTRODUCTION

Metastable austenitic steels are used in the structure of high field of superconducting magnets. In this application the alloys sustain high stresses in high magnetic fields at 4.2 K. Safe and effective design demands complete characterization of tensile, fatigue, and fracture toughness behavior of these alloys under the anticipated operating conditions. The characterization problem is complex because these metastable austenitic steels can undergo strain-induced martensitic transformation at cryogenic temperatures.[1,2] The available data on the behavior of metastable austenitic steels in magnetic fields at 4.2 K is somewhat confusing. It is known that the presence of a high magnetic field during deformation can enhance martensitic transformation in Ni, Ni-Cr, and Mn steels.[3-11] However, Goldfarb et al.[12] reported no detectable field-induced transformation in 304 at 4.2 K with deformation in a 3.7 T applied field. The tensile behavior[13-15] and the fatigue behavior[16] of 304L and 304LN, which are, respectively, less stable and more stable than 304, have also been investigated in some detail in higher strength fields. Yield strength and fatigue property changes between tests performed in a magnetic field and without a magnetic field were not significant while the ultimate strength increased only slightly. However, very recent work suggests that there may be a significant decrease in the fracture toughness of 304 at 4.2 K in a high magnetic field.[17] If true, this effect would necessarily impact the design of high field magnet structures. Hence, the present work was undertaken to determine the extent and nature of the fracture toughness change of 304 stainless steel at 4.2 K in an 8 T magnetic field.

EXPERIMENTAL

The composition of the 304 plate used for this experiment was 18.8Cr-8.0Ni-2.0Mn-0.4Si-0.04C-0.093N-0.027P-0.001S. J_{Ic} tests were performed in an 8 T magnetic field at 4.2 K in the bore of a superconducting solenoid. A single specimen compliance technique, in which load line displacement measurements are made using a clip gage, was used to determine the specimen compliance and thereby crack length. The specimens were all precracked at room temperature to a nominal a/w of 0.6. Testing and analysis was according to ASTM813-87.

Since it is possible that the presence of the field can alter the response of the load cell and clip gage used in these tests, care was taken to calibrate the response of these instruments in the actual test conditions. The clip gage response was measured for an imposed physical displacement in an 8 T magnetic field at 4.2 K. Figure 1 shows the clip gage response with and without an applied magnetic field. The gage response is linear in both cases but a change in slope occurs on application of the 8 T field. The load cell response at room temperature in the presence of the magnetic field was measured by placing the tensile machine under constant stroke control and moving the crosshead, and thus the load cell, from approximately 2.5 m from the top of the bore of the solenoid to its operating position of approximately 1 m above the top of the bore. Any magnetic field influence on the output of the load cell would be observable on traversing the magnetic field gradient. The effects of the magnetic field on the load cell and the stroke LVDT (linear variable differential transformer) displacement transducer were negligible.

Metallographic specimens were electroetched in a saturated aqueous solution of NaOH at 6 V to reveal grain boundaries and the martensite transformation products. The specimens were also coated with ferrofluid to decorate the ferromagnetic martensite phase. Permeability and metallographic measurements were used to qualitatively and quantitatively determine the extent of the martensitic transformation. The α' distribution of the transformation zone was determined by areal fraction measurements on high magnification optical micrographs. X-ray diffraction was performed on polished and slightly etched specimens and on the fracture surfaces.

RESULTS

The 4.2 K fracture toughness of the specimens tested in the 8 T magnetic field are compared with those of specimens tested without an applied field in Figure 2. The yield strength of the 304 plate at 4.2 K is 630 MPa (91 ksi) in the longitudinal direction and 600 MPa (87 ksi) in the long transverse direction. The spread in the fracture toughness data for each condition is extremely small, with almost no scatter in the 8 T, T-L orientation specimens. The in-field fracture toughness is significantly higher than that of the zero-field fracture toughness, in both the T-L and L-T orientations (30% higher for the T-L case and 40% higher for the L-T case).

Permeability measurements show that the high permeability α' phase forms near the fracture surface. The amount decreases with distance from the fracture surface. Metallographic examination of fracture surface profiles and x-ray diffractometry confirm this observation (Fig. 3). Beyond the observed plastic zone, α' was not found. The amount and extent of α' formation was qualitatively larger for the specimens tested in 8 T than for those tested in 0 T. The martensite formed in both cases had a fine lath morphology.

The relative distributions of α' in the transformation zone, as determined by metallographic measurements, are indicated in Figures 4. In both cases, the measured volume fraction of α' at a given distance from the crack plane was higher in the samples tested in the 8 T field than in the samples tested in zero field. Comparison of relative x-ray intensities of the α' 200 and 211 peaks with calculated relative intensities indicates a preferred α' texture. However, there was also a slight texture in the original austenitic material, so it is difficult to determine the degree to which α' texture was induced by the 8 T field.

Fractographic examination shows a mixed mode fracture surface that is typical for 304 broken at 4.2 K; the fracture mode is primarily ductile with some secondary cracking and some relatively planar areas associated with α' (Figure 5). These planar features are similar to those reported by Reed et al.[25] and by Tobler and Meyn[26], the latter identified them to be (111) planes of the γ phase. The fracture surfaces of the specimens tested in 8 T and those tested in 0 T are qualitatively similar and have similar areal fractions of planar patches associated with α' (25% and 30% respectively). All specimens show some secondary cracking parallel to the plane of the CT specimens and perpendicular to the primary crack plane (Figure 6). Some splitting was also observed on the planar areas with the secondary crack planes parallel to the specimen face and perpendicular to the primary crack plane. These cracks are parallel to α'-γ laminate planes.

DISCUSSION

304 stainless steels are metastable, tending to transform from an fcc structure to a more stable bct martensite at low temperatures. Low temperature, mechanical stress, and the presence of a magnetic field will increase the driving force for transformation. Mechanical deformation may also facilitate transformation by promoting the nucleation of the martensite phase. Thus the presence of these conditions would tend to shift the transformation temperature upward and enhance the extent of the transformation to α'.

The effect of pulsed[3-6,8,9,13] and steady[7,10] magnetic fields on the martensitic transformation behavior of steels of various composition has been examined. These studies indicated that the amount of transformation was enhanced in the presence of a field. The extent of transformation was shown to be a function of field strength and independent of frequency and the number of applied pulses.[6] The presence of an applied field during transformation was also found to have an influence on the orientation distribution of the martensite which was formed.[19,20] Satyanarayan[18] suggested that the amount of magnetically-induced α' is proportional to the product of the field strength and saturation magnetization difference between the parent γ and the product α' phase. Fultz et al. estimated the increase in transformation temperature to be approximately 2 degrees K per Tesla.[13]

During crack initiation and propagation, there is local deformation ahead of the crack tip due to the presence of a stress concentration. Therefore, a region exists ahead of the crack tip which has a higher transformation temperature than the bulk of the specimen and thus has a greater tendency to transform into α'. This tendency would be increased further by the presence of a magnetic field. This excess transformation product, above that expected from strain-induced transformation alone, and the timing of this transformation are expected to have an effect on fracture behavior.

Some probable effects of the transformation on the fracture toughness include: magnetostatic forces acting to reduce crack tip stress intensity, crack "shielding" due to dilatational and deviatoric components of the transformation strain acting to reduce the effective applied stress intensity on the crack tip, strain hardening rate increase due to the formation of α', and crack deflection and secondary cracking due to the presence of the harder α' phase in the γ matrix. These mechanisms are all expected to be operating to some degree to increase the measured fracture toughness. Possible factors which act to reduce fracture toughness when α' forms ahead of the crack tip, assuming no change in fracture mode, include increased effective stress intensity at the crack tip due to the transformation strain in the wedge-shape volume directly ahead of the crack tip bounded by angles of $\pm\pi/3$ from the crack plane[21] and the presence of α' containing regions of lower resistance to crack initiation and propagation at cryogenic temperatures.

Magnetostatic forces are of some significance to these measurements. The volume of ferromagnetic α' formed during deformation in a magnetic field has a direct effect on the crack tip stress intensity. Since magnetostatic energy is minimized by containing the magnetic flux within the ferromagnetic material, the plates would tend to resist separation, with forces that can be calculated. During J_{Ic} tests on CT specimens in a solenoidal field, with the loading direction parallel to the solenoid axis, the formation of martensite around the crack would create a closing load on the crack. Since the tests were conducted under displacement control, there would be a higher recorded load per unit crack opening and advance in the magnetic field. This effect should also increase with crack extension since the martensite containing volume also increases. Calculations indicate that the magnitude of this effect accounts for approximately a 1 MPa-$m^{1/2}$ reduction in crack tip stress intensity, representing a 3% improvement in T-L and a 2% improvement in L-T measured fracture toughness.[31]

It is known that martensitic transformation, when properly situated both spatially and temporally with respect to the crack, can shield the crack from a portion of the applied stress, causing an apparent toughening of the material.[21,22] 304 is thermally and magnetically stable unless plastic deformation at cryogenic temperatures occurs; then the martensite forms within the plastic zone of the crack. This zone is bounded by material which remains elastic and thus has not transformed. This configuration is seen in our specimens. The volume expansion due to martensite formation is therefore constrained by the elastic untransformed material, altering the stress field around the crack. If the martensite is behind a front bounded by angles of $\pm\pi/3$ with respect to the crack plane, the dilatation will impose a compressive stress, thus decreasing the effective stress intensity at the crack tip. The measured fracture toughness would thus increase. If transformation occurs in the wedge ahead of the crack tip bounded by the front, the dilatation results in a tensile stress on the crack tip, thus increasing effective stress intensity and decreasing measured fracture toughness. This stress intensity change is a function of the transformation strain, transformation zone size and geometry, and volume fraction transformed. A two-dimensional calculation[31] assuming plane strain conditions, a transformation zone and martensite distributions similar to that observed, and accounting only for the dilatational component of the transformation strain, yields a difference in stress intensity between the 0 T and the 8 T case of approximately 2.3 $MPa\text{-}m^{1/2}$. The calculation, however, assumes isotropic dilatational effects only. Other assumptions, such as accounting for the transformation shear strains, and the preferred orientation of the transformed particles can increase the difference in effective stress intensity significantly.[23]

Comparison of the tensile deformation behavior of 304L and 304LN stainless steels in high strength magnetic fields and without applied magnetic field at 4.2 K indicates that there is no detectable change in the yield strength between the two cases, but there are differences in the post yield deformation behavior due to transformation to α'.[14] In a magnetic field, the specimens show a slight reduction in flow stress right after yield and a higher strain hardening rate at higher strains than specimens tested without a magnetic field. Fultz and Morris[14] found that over most of the range of uniform elongation both 304L and 304LN had higher strain hardening rates when tested in an applied field. This difference increased with increasing true strain. The increase in strain hardening rate with applied field would thus allow the transformed volume of material to absorb more energy for a given imposed strain in the field. Fracture toughness has been modelled as a function of various material microstructural and tensile parameters.[24,30,32,33] For ductile mode fractures, fracture toughness varies, either implicitly or explicitly, with the strain hardening rate of the material. Fultz, et al.[14] show maximum strain hardening rate differences of 10% and 17% between the 0 T and the 18 T cases for 304LN and 304L respectively at 77 K. If the differences for 304 at 4.2 K are comparable, then we should expect a similar difference in the fracture toughness. Precise quantitative estimates, however, are difficult to make because of the geometry of the transformation zone and the nonuniform distribution of martensite therein. But it appears reasonable to assume that the fracture toughness differences would scale with strain hardening rate differences and thus the magnitude of the fracture toughness differences would be on the same order as the strain hardening rate differences.

Our results showing an improved fracture toughness in a magnetic field do not agree with those reported by Fukushima et al.[17] They reported a decrease in J_{Ic} with application of a magnetic field and suggested that their results can be explained in terms of a theory developed by Shindo[28,29], which predicts an increase in crack tip stress intensity for a crack in a magnetically soft elastic body in a magnetic field oriented normal to the crack plane. However, the assumptions used in deriving the theory do not correspond to the experimental conditions in this case. The ferromagnetic phase in the test specimens develops in the presence of the magnetic field during deformation and is localized to an area around the crack. It is thus nonuniformly distributed contrary to the uniform body assumed in the theory. Furthermore, the theory was developed for small field strengths and assumed a linear increase in magnetic susceptibility with field. Thus it is not obvious that the predicted stress intensity increase should be seen with our experimental conditions.

At least part of the disparity between Fukushima's data and the current data may be due to differences in test procedures or in the stability of the alloys. The stability and thus the mechanical properties of 304 is sensitive to factors such as C, N, and Ni content[27] and grain size. The alloy used in this experiment had a slightly lower C and Ni content and thus may be less stable than that used by Fukushima et al.[17] (N content was not reported). Hence, the mechanical properties affected by the presence of α' can be different. Although the reasons for this difference in fracture behavior cannot be stated with certainty, it appears that the measured fracture toughness may be sensitive to test procedures and alloy chemistry and microstructure.

Although the mechanisms enumerated above all serve to increase effective fracture toughness, they do not fully account for the measured differences (30-40%). Other mechanism, such as the shear component of the transformation strain, have not been considered in detail here and may help better explain the measured differences.

CONCLUSIONS

The fracture toughness of 304 stainless steels tested in this experiment shows an improvement in an 8 T field at 4.2 K. This improvement is expected as a result of magnetostatic effects and transformation strain differences due to the excess martensite formed within the magnetic field, and the increase in strain hardening rates. However, the mechanisms discussed cannot completely explain the observed fracture toughness enhancement. Further examination of the fracture behavior of these alloys over a wider microstructure and chemistry range and under different loading conditions is also be warranted. Further theoretical work would also be useful. In particular, a theory which deals with the effects of a growing martensite transformation zone around a crack, including the effects of the transformation shear strains, in a strong magnetic field would be of interest. A comparison of fully pretransformed and fully stable specimens may also be of interest for isolating the magnetostatic effects.

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Figure Captions

Figure 1. Displacement gauge response at 4.2 K in 0 T and 8 T magnetic fields.

Figure 2. Fracture toughness variation of 304 stainless steel at 4.2 K with magnetic field and with orientation. Error bars represent the spread in the observed fracture toughness.

Figure 3. Optical fracture profiles showing the approximate extent of the α' zones for T-L (left) and L-T (right) orientations at 4.2 K for the two magnetic field conditions. Dark elongated features are sulfide stringers.

Figure 4. The distribution of α' in the transformation zone of T-L (left) and L-T (right) compact specimens tested at 4.2 K in 8 T and 0 T fields.

Figure 5. SEM micrograph of a typical planar fracture surface feature associated with α'-γ delamination.

Figure 6. SEM micrograph of an L-T specimen tested at 4.2 K showing secondary cracking associated with primary crack growth. Primary crack advance is from left to right. These secondary cracks are observed in all tested specimens.

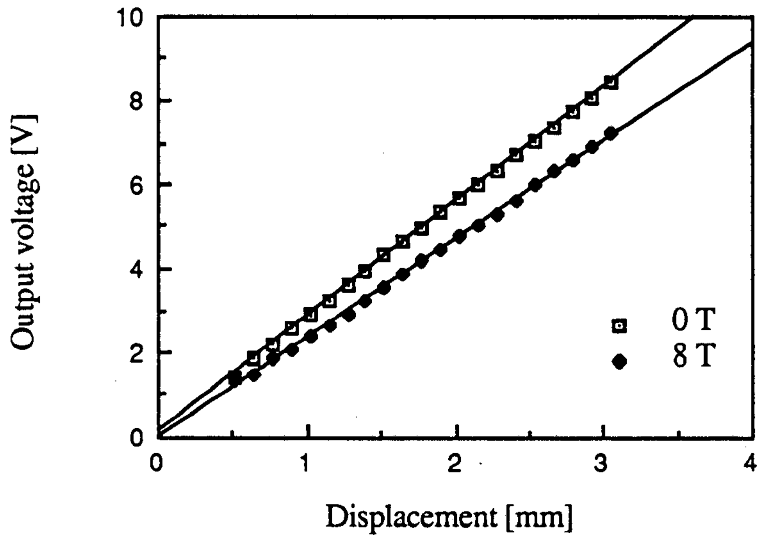

Figure 1

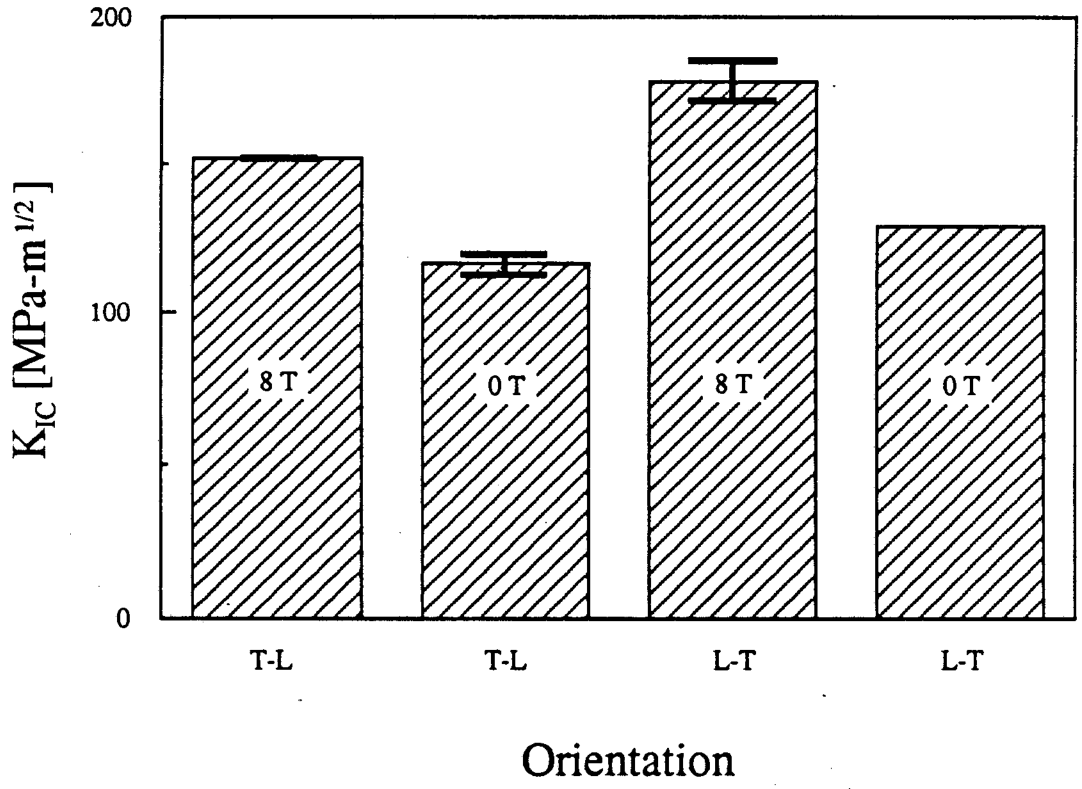

Figure 2

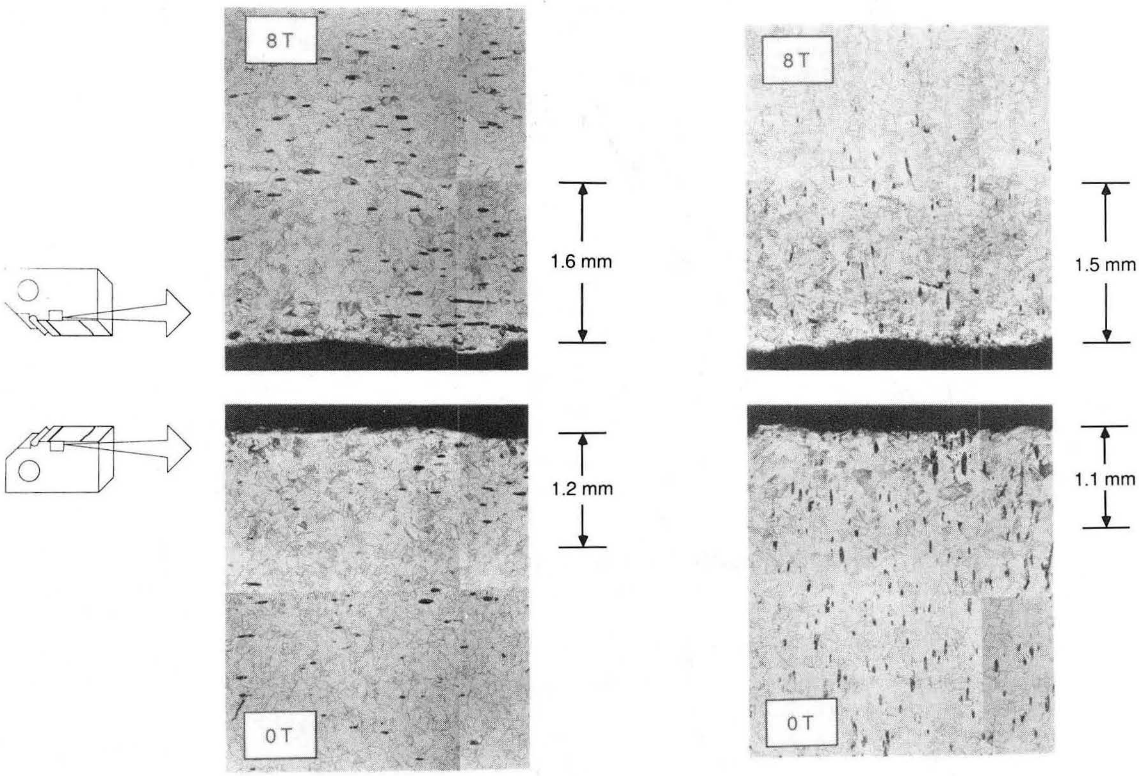

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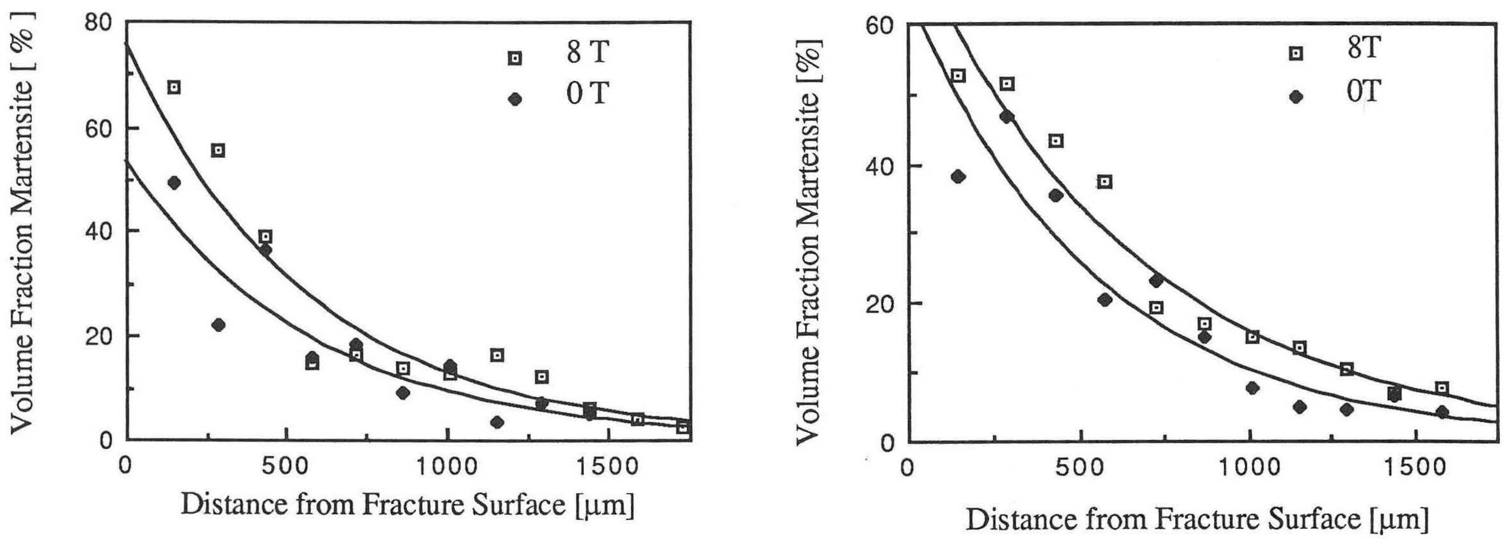
Figure 3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20
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1000
1500
Distance from Fracture Surface [μm]
8 T
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40
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0
0
500
1000
1500
Distance from Fracture Surface [μm]
8T
0T

Figure 4

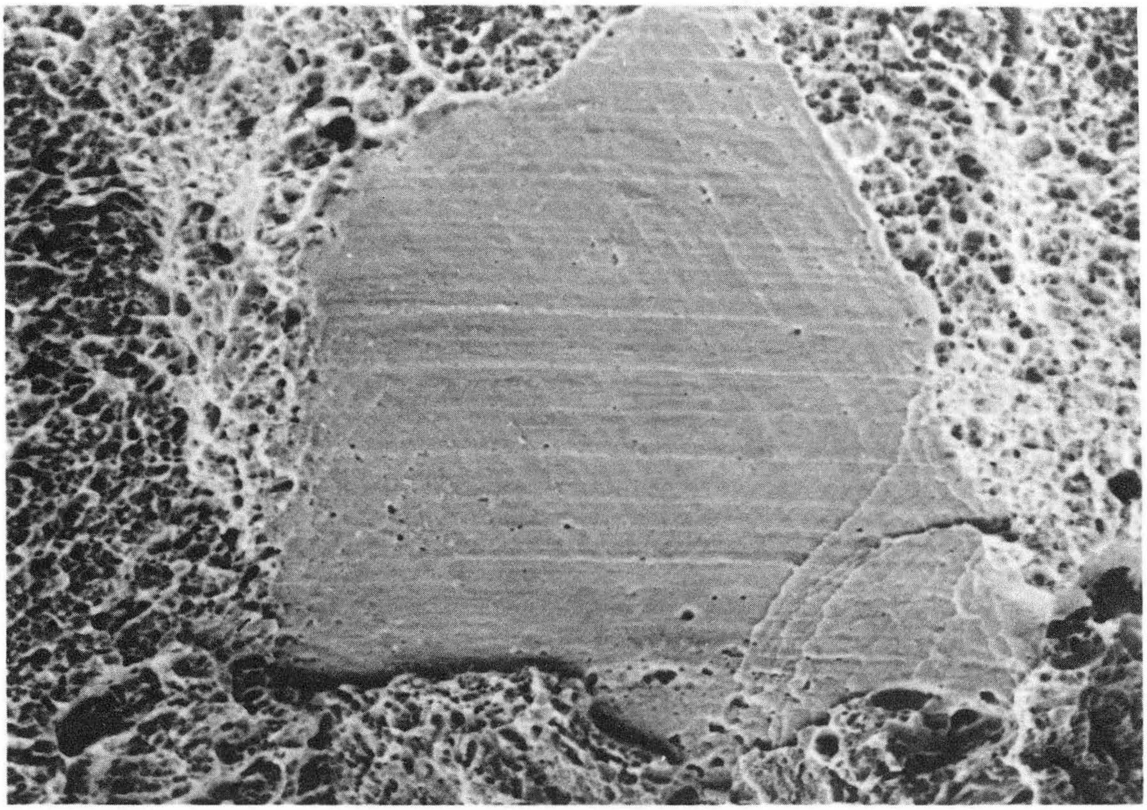

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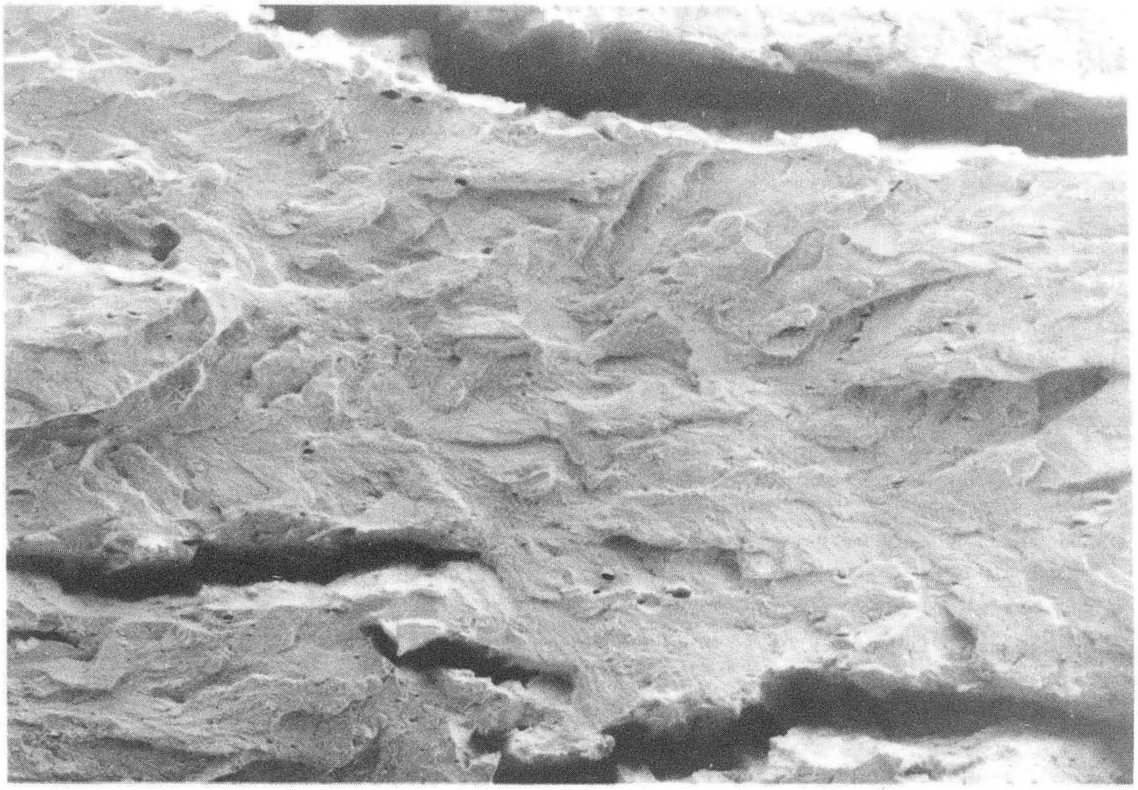
Figure 5



XBB 880-9769 150 μm

Figure 6

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