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Lawrence Berkeley Laboratory
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Materials & Molecular Research Division

Presented at the 3rd International Al-Li Conference, Oxford, England, July 9-11, 1985

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June 1985

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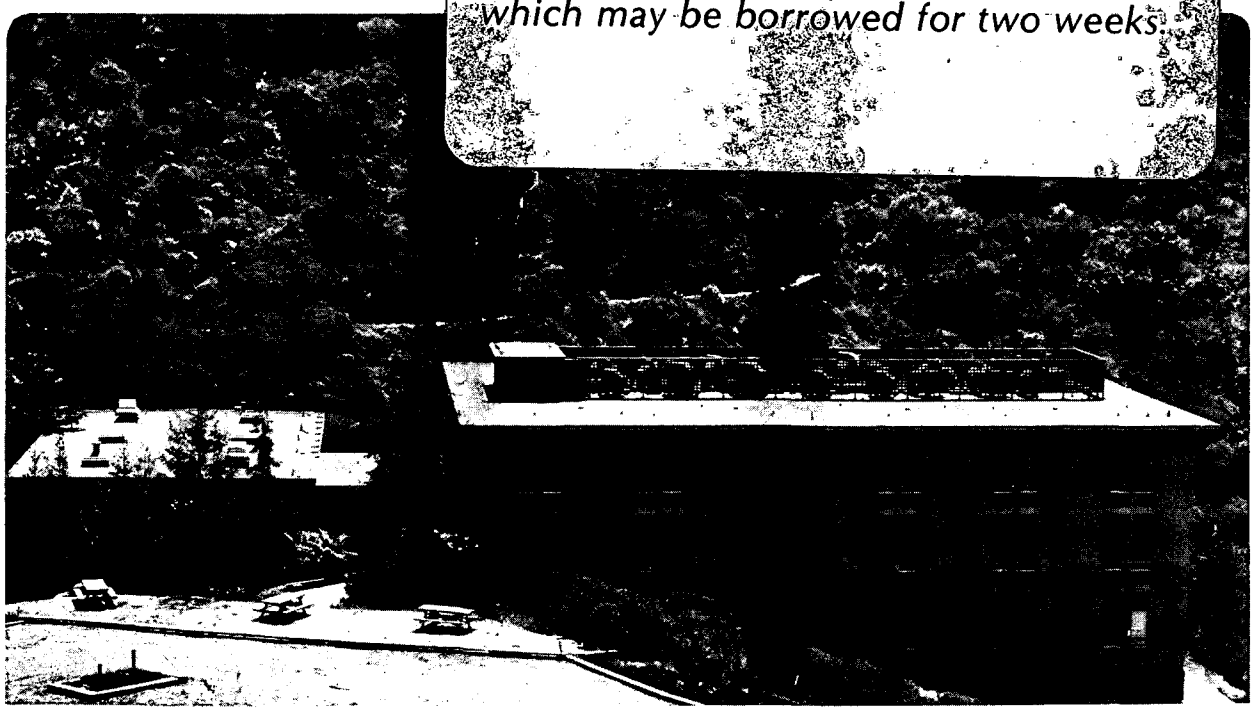

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Prepared for the U.S. Department of Energy under Contract DE-AC03-76SF00098

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THERMOMECHANICAL PROCESSING OF A TWO-PHASE AL-CU-LI-ZR ALLOY

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SUMMARY

The effect of large second-phase particles on the workability, mechanical properties, and response to thermomechanical processing of an Al-2.49Cu-2.37Li-0.13Zr alloy was investigated. The solution-treated alloy was compared with a dispersoid-containing alloy fabricated by aging the base alloy at an intermediate temperature to bring out second-phase particles. Extensive deformation and short heat treatments that cause some of the solute to be incorporated into dispersoid particles were both found to be beneficial to the mechanical properties of the alloy. In specimens which failed intergranularly, the presence of dispersoid particles did not significantly affect the fracture properties.

I. INTRODUCTION

Thermomechanical treatments of aluminum alloys are usually designed to avoid precipitation of large incoherent second-phase particles. These particles, which often appear in heavily alloyed materials, are generally thought to be harmful to the mechanical properties of the alloy [1,2]. However, the effect of second phase particles is alloy dependent. The properties of these second-phases and their effects on the microstructure, workability, and response to thermomechanical processing have not yet been characterized for aluminum-lithium alloys. In this study we have investigated the effect of various thermomechanical treatments on the microstructure and mechanical properties of an Al-Cu-Li-Zr alloy.

A number of incoherent phases exist in the aluminum-copper-lithium phase diagram. This study focussed on the T_2 phase (Al_5CuLi_3, cubic structure, isomorphous to $Mg_{32}(ZnAl)_{49}$ [3]). To facilitate comparison between alloys with and without the T_2 phase, an Al-2.49Cu-2.37Li-0.13Zr alloy was selected which could either be solutionized or processed to bring out the T_2 phase as large incoherent particles.

II. EXPERIMENTAL PROCEDURE

The specimens used in this study were fabricated from an ingot cast at the Alcoa Technical Center. Prior to receipt, the ingot was

homogenized, hot rolled to one half inch plate and then solution treated and stretched 2%. The composition of the alloy was Al-2.49Cu-2.37Li-0.13Zr-0.06Fe-0.03Si-0.02Ni-0.0008Na-0.0007Ca in weight percent. The general appearance of the as-received material is shown in Figure 1. The grains are elongated as a result of hot-rolling and unrecrystallized due to the presence of Zr in the alloy.

Six thermomechanical processing sequences were considered. The processing schedules are summarized in Table 1.

The dispersoid-containing alloy (450 in Table 1) was prepared by heat treating for four hours at 450°C to precipitate the T_2 phase. This heat treatment temperature was chosen on the basis of a linear extrapolation from isothermal sections of the Al-Cu-Li phase diagrams given by Mondolfo [3] and Hardy and Silcock [4]. However, x-ray diffraction results indicate that the T_1 phase and trace amounts of δ phase were also present.

To study the effect of these incoherent phases on the response of the material to thermomechanical processing, both alloys were extensively cold-worked. Extensive cold-forming without annealing (>150% reduction) caused the alloy to fail in shear, probably due to the fact that the Zr in the alloy prevented easy dynamic recrystallization. This problem was particularly pronounced in the dispersoid alloy. However, it was possible to design a processing sequence which allowed extensive deformation. A form rolling process in which single large deformation steps were alternated with 1 hour anneals at the "solutionizing temperature" (520°C and 450°C, respectively) was used to deform the specimens a total of nearly 1000% in six steps. In the last step the material was swaged into rod. The rod axis corresponds to the rolling direction of the original plate. The grain structure of the deformed material is illustrated in Figure 2. After the final anneal, both alloys (Base-D and 450-D) were stretched 2% before final aging.

For comparison, additional specimens (Base-A and 450-A) were prepared that received the same heat treatment as the deformed specimens above, but were not deformed.

Rockwell B hardness was used to determine the peak-aged condition for each processing sequence. All specimens were peak-aged at 190°C before mechanical testing, with the exception of the deformed base alloy which was tested in a slightly overaged condition.

The mechanical properties of the specimens were compared in the peak-aged condition using subsize tensile and notched tensile bars machined in the longitudinal direction. The specimens were similar in design to those used by Rack and Edstrom [5], with a 3.2mm gauge diameter and a 25.4mm gauge length. The notches were 60° with a root radius of less than 50μm. All tests were conducted using at a strain rate of 1.7×10^{-4} sec^{-1}.

Microstructural examination was carried out using a combination of techniques. Polished sections were examined by optical and scanning electron microscopy after etching by Keller's etch (for grain size) and 1% HF (for second-phase particles) [4]. Samples were prepared

for transmission electron microscopy by jet polishing at 12V, -30°C in a 1:3 mixture by volume of 70% nitric in methanol. The foils were then examined in a Philips 301 at an operating voltage of 100kV.

III. RESULTS

1. Microstructure.

a. Base Alloy

The base alloy provided a benchmark for comparison with other alloys. Before aging, optical microscopy showed the material to be relatively free of large inclusions and second-phase particles. At peak strength the alloy was hardened by δ', T_1' and T_2' phases.

The main difference between the deformed base alloy (Base-D) and the undeformed alloy was the smaller and more elongated grains in the deformed alloy. The subgrain size in the deformed alloy was also substantially smaller than in the undeformed alloy.

b. Dispersoid Alloy

An optical micrograph of the alloy after aging at 450°C for 4 hours is shown in Figure 3. A number of particles are visible. Transmission electron microscopy (Figure 4) indicates that several different types of particles exist, including T_1, T_2 and Al_3Zr. The particles are relatively uniformly dispersed and do not tend to be concentrated along grain or subgrain boundaries. The subgrain structure, similar to that observed in the base alloy, is shown in Figure 5.

The particles in the deformed dispersoid alloy (450-D) are almost the same size as those in the undeformed alloy despite the additional six hours at 450°C during the deformation process. It is not clear whether the particles deformed or fractured during deformation, or if their growth was inhibited. Preliminary transmission electron microscopy results indicate that the particles tend to be aligned so that they present a large cross section in planes parallel to the rod axis and a small cross section perpendicular to it. Figure 6 illustrates that as a result of the extensive deformation the subgrain size in the deformed alloy was again much smaller than in the undeformed alloy.

2. Mechanical Properties

The mechanical properties of all six samples are shown in Table 2. All processings result in very low elongations. The ratio of notched tensile strength to yield strength (NYR) is used as a measure of toughness in these small samples. The notch yield ratio is plotted against yield strength in Figure 7. The strength-toughness characteristic of some commercial 2XXX and 7XXX alloys is shown for comparison [6]. As noted by Feng et al. [7] for alloys similar to 2020, the tensile elongation and fracture toughness do not correlate well.

As shown in Figure 7a, the dispersoid and base alloys have

similar mechanical properties in spite of the fact that their microstructures are quite different. Perhaps surprisingly, the dispersoid alloy seems to ride a better strength-toughness characteristic. However, after an additional six hours at 450°C, its properties drop off significantly. The deformed disperoid alloy (450-D), which received the same heat treatment as the 450-A specimen, shows significantly better properties.

As expected, the properties of the solutionized base alloy, plotted in Figure 7b, were much less sensitive to thermomechanical processing of this kind. The deformation processing improves the strength-toughness relation slightly for the base alloy. The Base-A treatment results in both lower strength and higher toughness.

The notch yield ratio has been found to correlate well with plane stress fracture toughness in sheet specimens of alloys of similar composition and mechanical properties to those tested here [8]. Although the stress-state varies among these samples, the trends indicated by the NYR are probably valid over relatively small ranges of strength. The apparent toughness of the low strength samples is almost certainly spuriously high with respect to the other specimens since these specimens are nearer plane stress. Nevertheless, the deformed dispersoid specimen (450-D) is clearly considerably tougher than the other samples.

3. Fracture Behavior

a. Undeformed alloys.

The notched tensile fracture modes of the undeformed alloys are shown in Figure 8. The macroscopic failure mode was nearly identical for both undeformed alloys. Both exhibited significant longitudinal cracking. The longitudinal cracks are intergranular, as observed by Wert and Lumsden [9] for an Al-Li-Cu-Mg-Zr alloy. The microscopic failure modes were relatively similar although not identical. The base alloy exhibited considerable ductility. The fracture surface of the dispersoid alloy contains some ductile regions and some regions of transgranular shear. The smooth tensile specimens failed in shear at low elongations.

b. Deformed Alloys.

The notched tensile fracture of the deformed dispersoid alloy was completely ductile as shown in Figure 9. No indication of the grain or subgrain boundaries is present in the fracture surface. The deformed base alloy (Base-D) appears to be at a transition point with respect to fracture mode. Although the two specimens tested had almost identical yield strength and notch toughness, they exhibited two different fracture modes. One specimen (pictured in Figure 9) fractured in a ductile manner although there were some regions of intergranular fracture. The other specimen showed significant longitudinal intergranular cracking like that seen in the undeformed alloys in Figure 8 and did not show significant ductility. The smooth tensile specimens all failed in shear.

c. Annealed Alloys.

These specimens (Base-A and 450-A) both exhibited a macrosco-

pic fracture mode similar to the undeformed specimens (Base and 450). The fracture surface of the 450-A specimen is shown in Figure 10. Considerable longitudinal intergranular cracking was observed. Consistent with the lower strength of the annealed samples, these specimens appeared to be more ductile than the Base and 450 samples.

IV. DISCUSSION

The similarity of the mechanical properties and fracture appearance of the undeformed dispersoid and base alloys indicates that the particles present in the microstructure of the dispersoid alloy do not significantly affect the fracture properties of the alloy because the 'weak link' which leads to failure remains unchanged. As long as intergranular weakness controls the fracture mode, the material is relatively immune to inclusions in the matrix.

Deformation of the base alloy raises both the yield strength and the notched tensile to yield strength ratio slightly. However, deformation of the dispersoid alloy results in an entirely ductile fracture mode and high toughness, albeit at much lower strength. The low yield strength in this condition is probably a consequence of the large quantity of solute incorporated into the equilibrium phases precipitated at 450°C. The large volume fraction of second-phase particles leaves little solute to be precipitated as coherent strengthening phases. Once the alloy is weak enough to eliminate the laminar intergranular fracture mode seen in the undeformed alloys, the dispersion of precipitates in the matrix becomes important. An additional factor which may contribute to the more ductile behavior exhibited by the deformed specimens is the reduced grain size. The control specimen 450-A, which received the same heat treatment but was not deformed shows relatively poor strength and toughness. Its poor toughness confirms that the deformation process, as well as the precipitation of solute, plays a role in developing the good properties of the deformed dispersoid alloy.

V. CONCLUSIONS

The influence of the dispersoid phases present in this alloy may be summarized as follows. Large equilibrium phases are tolerable in relatively brittle alloys, and in small amounts may even have a beneficial effect on the mechanical properties of the alloy. Although these phases may make deformation more difficult, this problem can be overcome, at least for cold-forming. For alloys which fail intergranularly, precipitates in the bulk do not significantly influence the fracture mode or the mechanical properties. If other microstructural changes (e.g. change of grain size, lack of hardening precipitates) promote a ductile fracture mode, then the large inclusions will control the fracture mode. In this case the inclusions lower the strength-toughness relation. Finally, large equilibrium phases can incorporate a large fraction of the total solute in the alloy and thereby significantly reduce the hardenability of the alloy.

ACKNOWLEDGEMENTS

The authors are grateful to S.L. Verzasconi, Materials and Molecular Research Division, Lawrence Berkeley Laboratory for experimental assistance, to the Alcoa Technical Center for supplying the material used in these experiments and to W.H. Hunt, Jr., Alcoa Technical Center, for helpful advice. This work was supported by the Director, Office of Basic Energy Sciences, Materials Science Division of the U. S. Department of Energy under Contract No. DE-AC03-76SF00098. One of the authors, JG, was supported by an ATT Bell Laboratories Scholarship.

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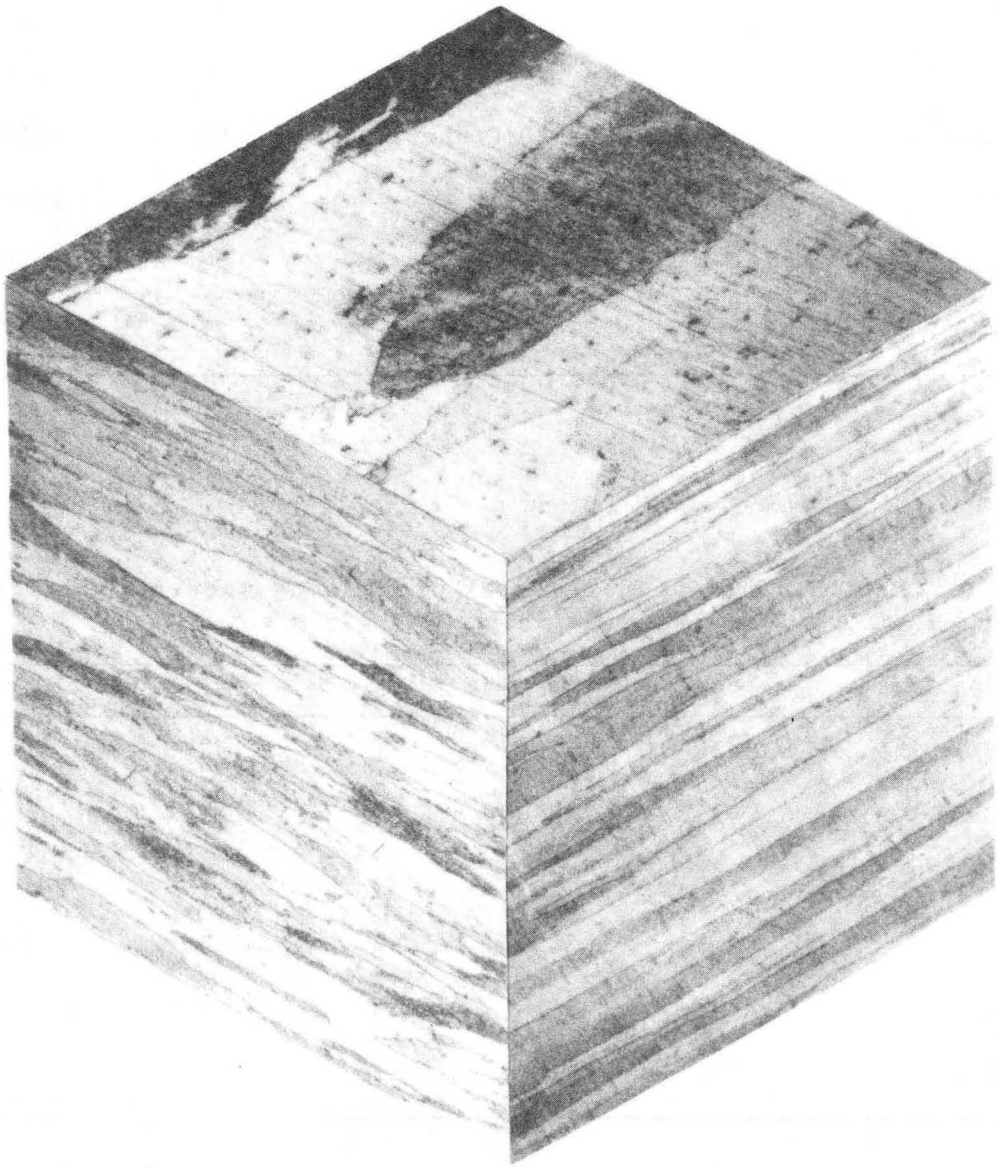
Table 1. Final thermomechanical treatments. The as-received condition is solution treated and stretched 2%.

Specimen	450°C	Form Roll/Anneal	Stretch	Peak-age
Base	--	--	--	190°C/8hrs
Base-D	--	1000% + 520°C/6hrs	2%	190°C/15hrs
Base-A	--	520°C/6hrs	2%	190°C/24hrs
450	4 hrs	--	2%	190°C/20hrs
450-D	4 hrs	1000% + 450°C/6hrs	2%	190°C/16hrs
450-A	4 hrs	450°C/6hrs	2%	190°C/20hrs

Table 2. Mechanical properties of processing sequences described in Table 1. All strengths are given in MPa.

Specimen	σ_{YS}	σ_{UTS}	σ_{NTS}	σ_{NTS}/σ_{YS}	Elongation %
Base	430	475	571	1.3	3.2
Base-D	453	486	646	1.4	3.0
Base-A	418	448	590	1.4	2.5
450	469	520	662	1.4	3.8
450-D	341	381	558	1.6	4.0
450-A	378	412	521	1.4	2.1



500 μm

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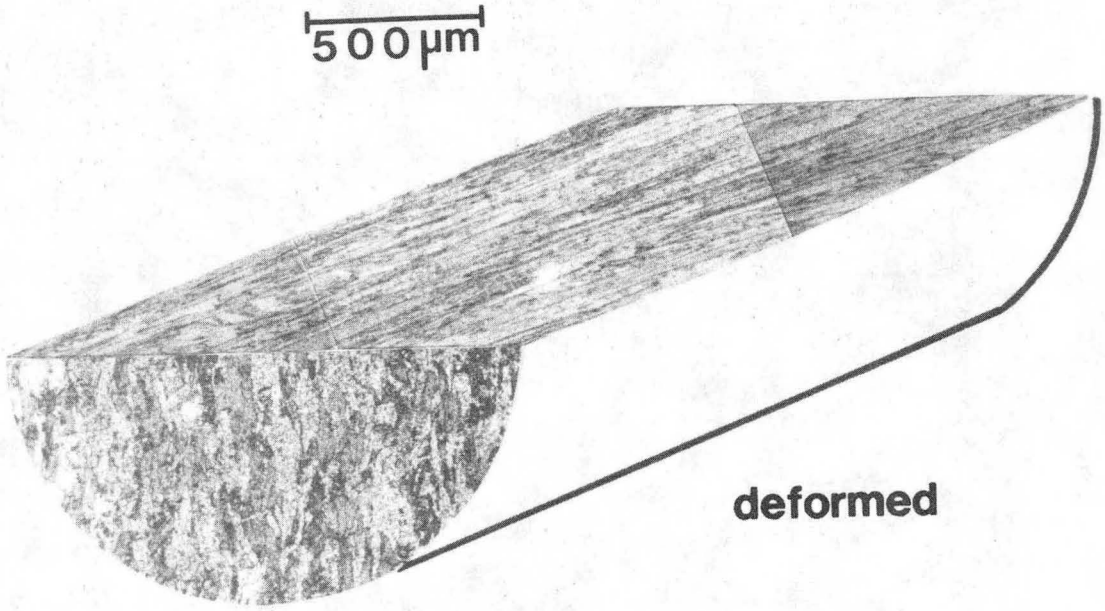
1. Optical micrograph of grain structure of as-received material.



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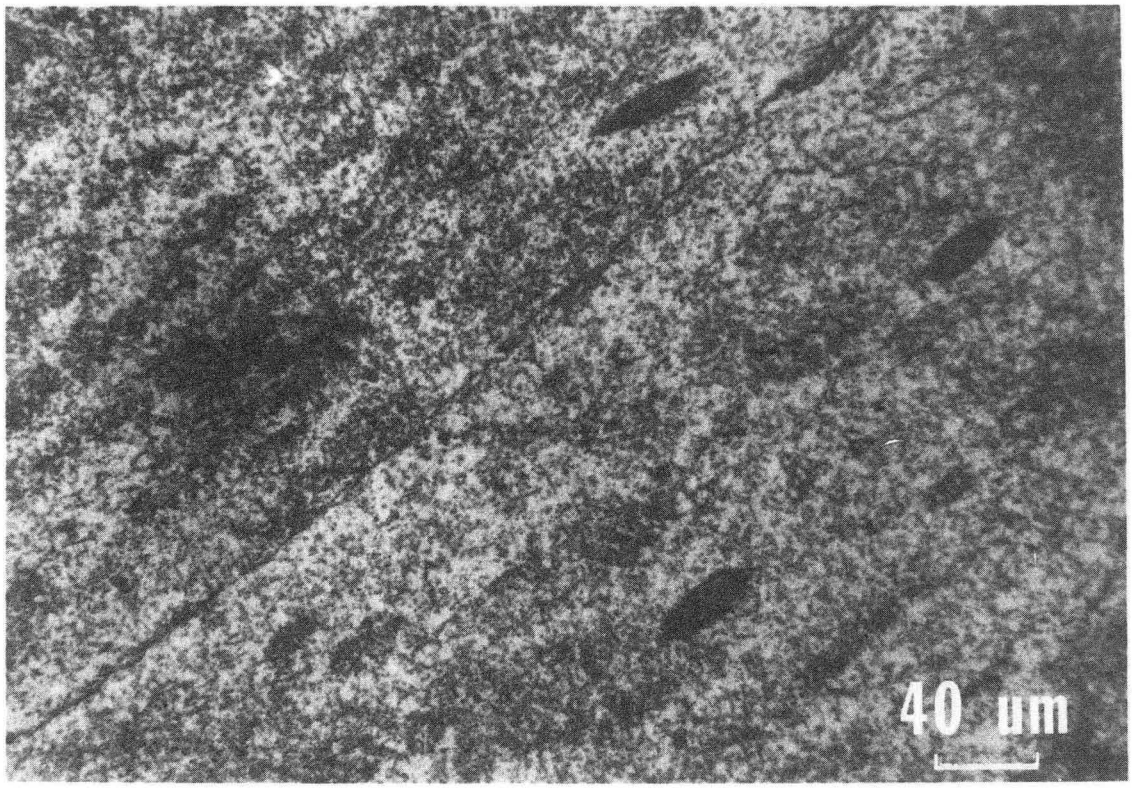
2. Optical micrograph of grain structure of alloy after 1000% deformation to rod.



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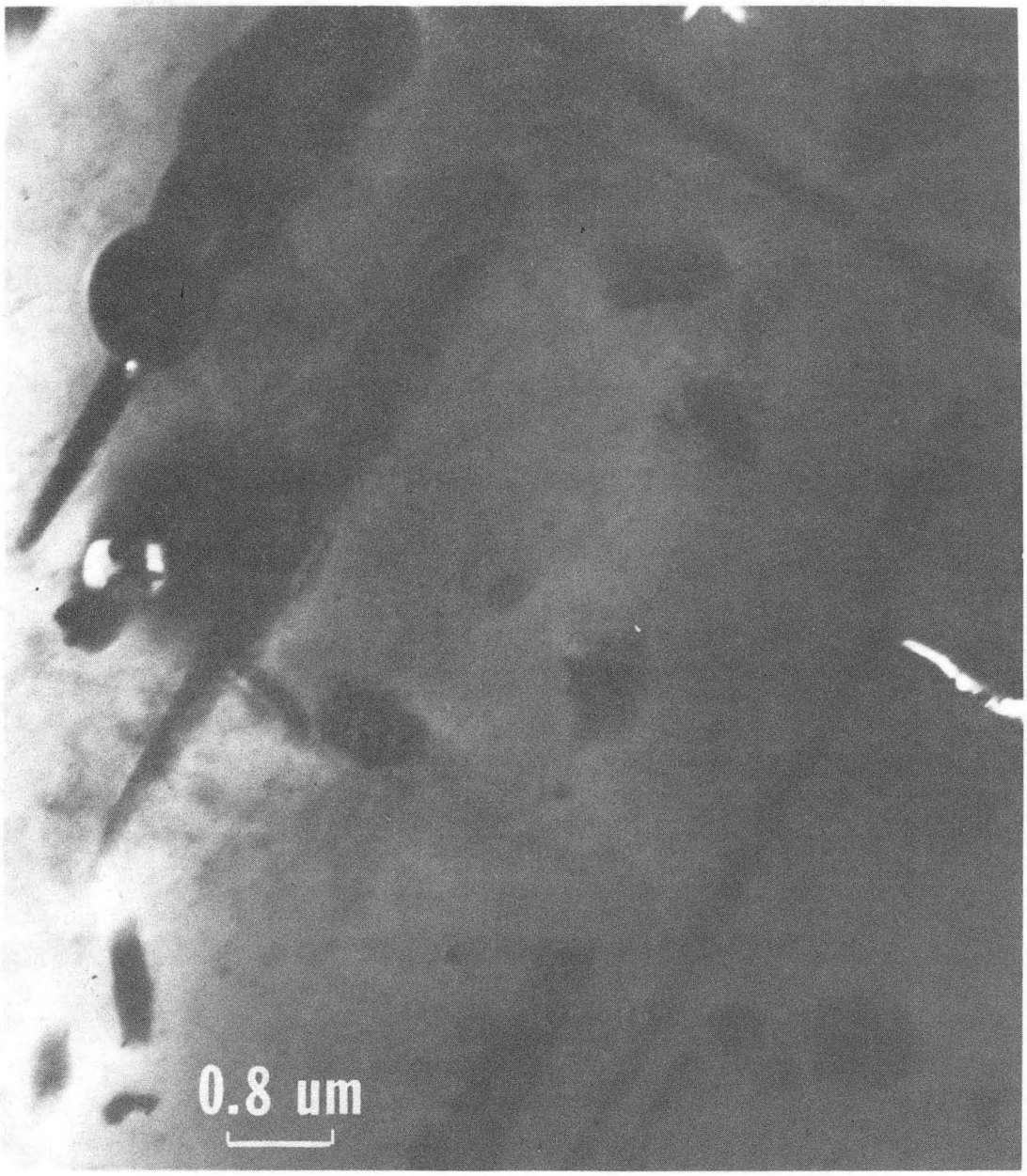
3. Optical micrograph of particles present in the alloy after 450°C/4hrs.



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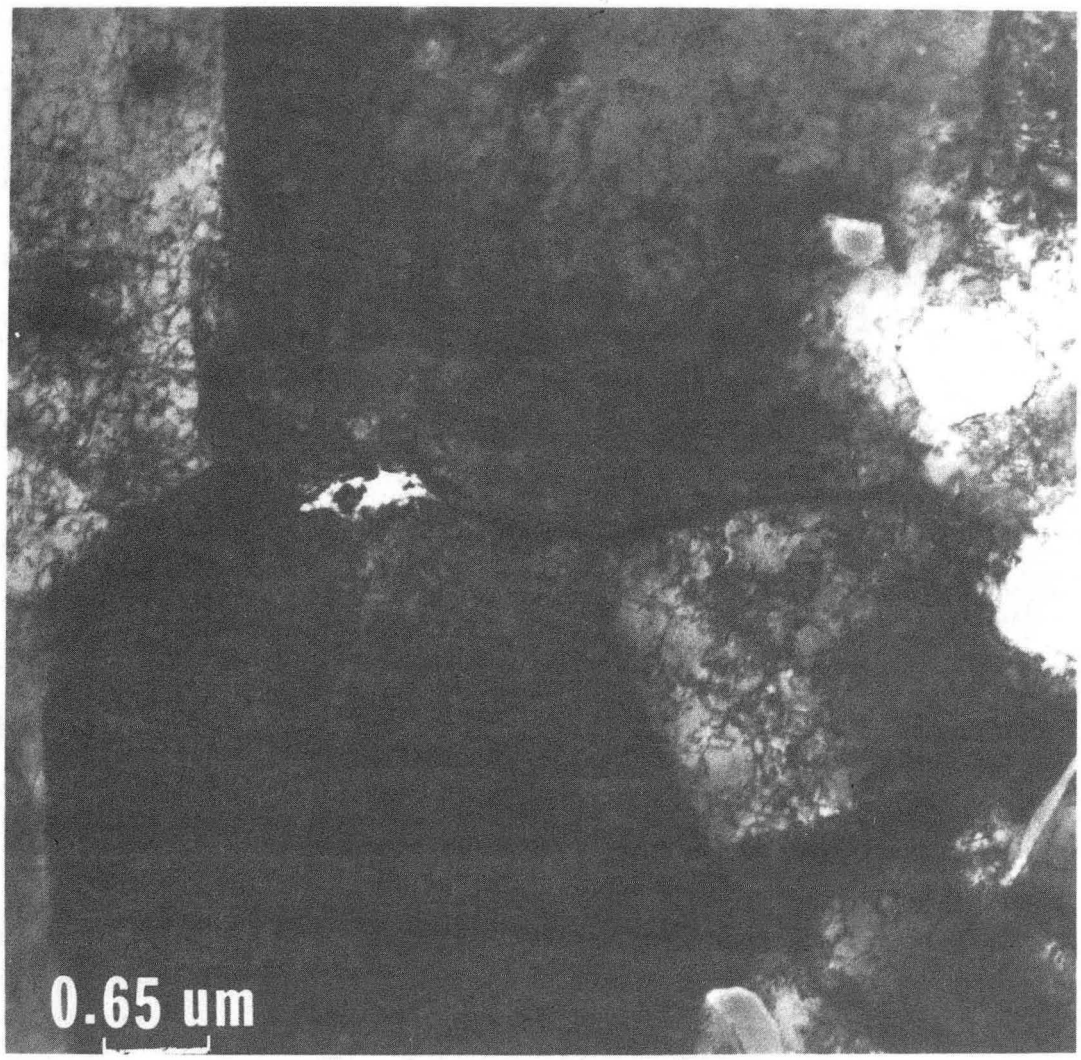
4. Transmission electron micrograph of particles present in the alloy after 450°C/4hrs.



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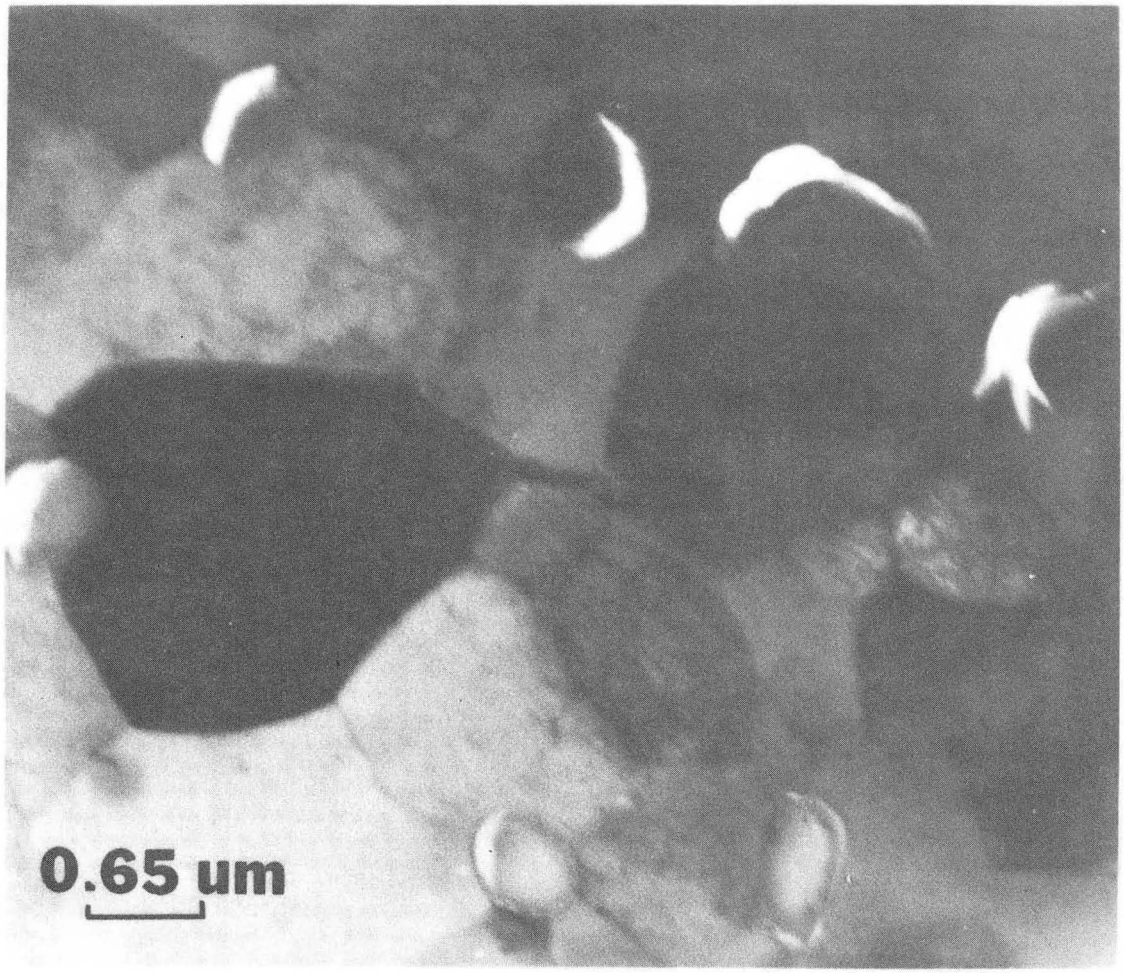
5. Transmission electron micrographs of the subgrain structure after 450°C/4hrs.



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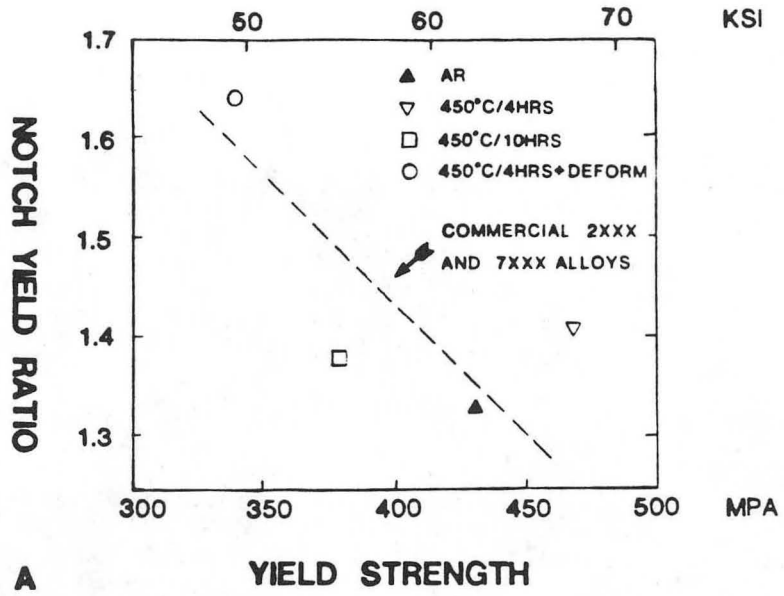
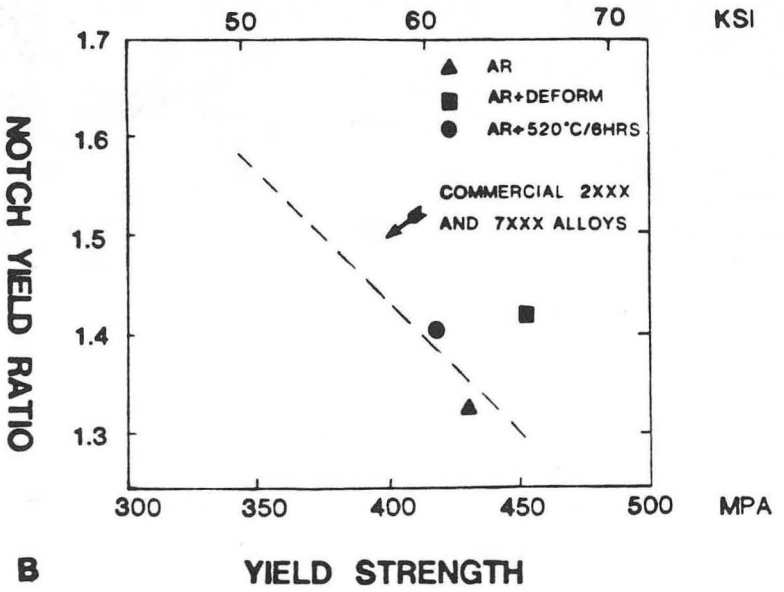
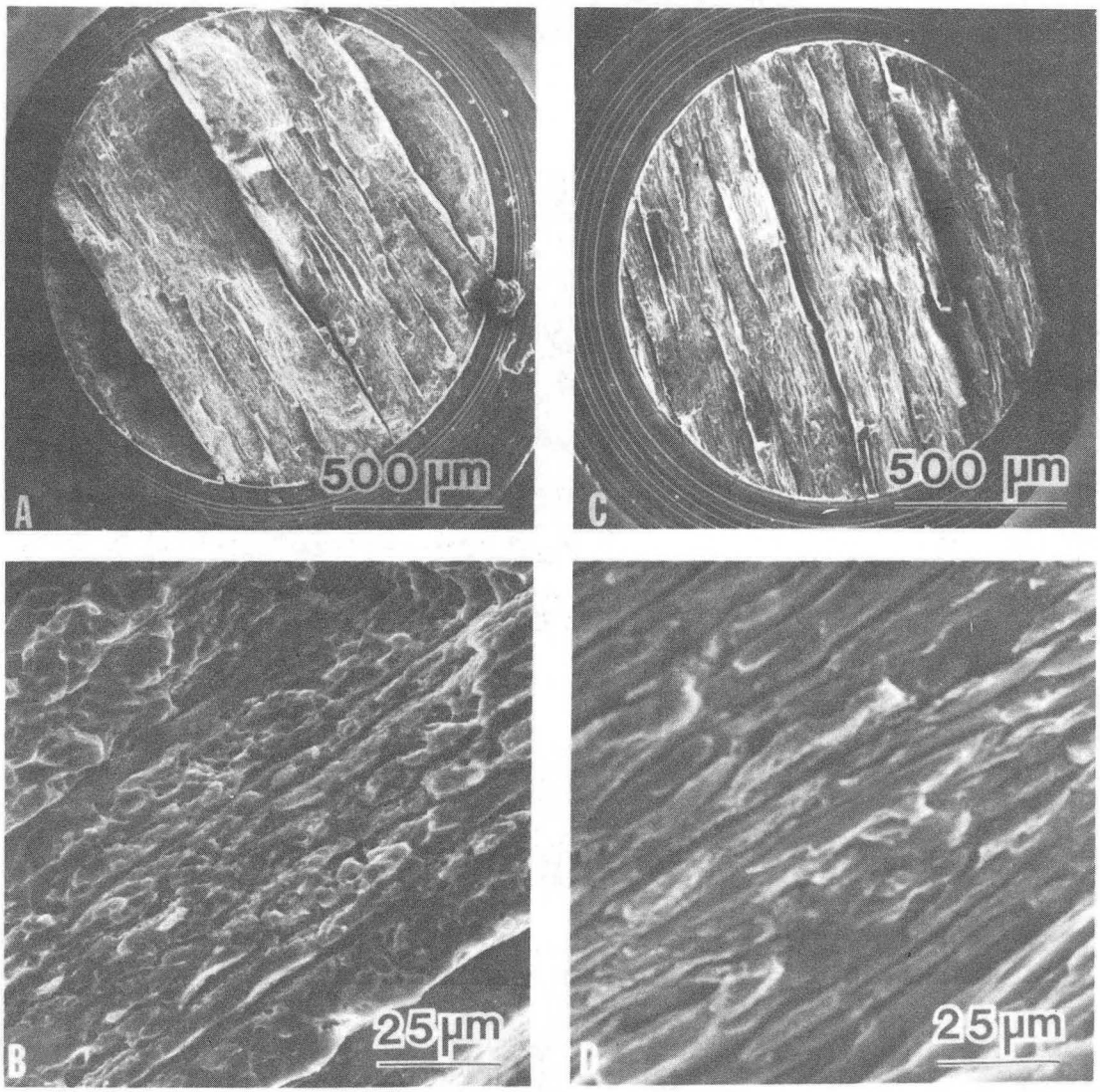
6. Transmission electron micrographs of the subgrain structure after 450°C/4hrs + deformation.





7. Notch yield ratio versus yield strength for (a) dispersoid-containing samples and (b) base alloy processings.



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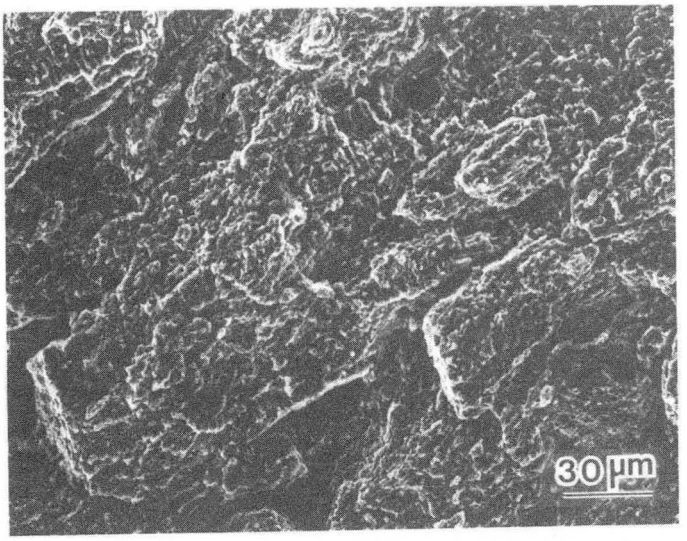
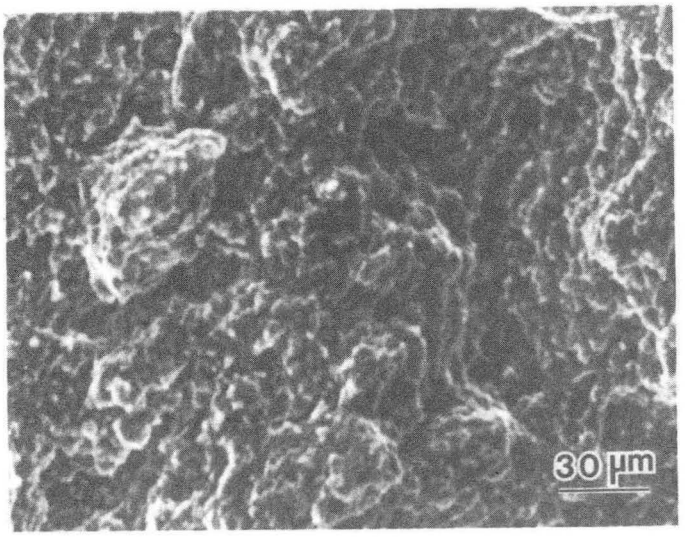
8. Notched tensile fracture modes of undeformed alloys. (a) and (b) base alloy, (c) and (d) dispersoid alloy.

DEFORMED SPECIMENS

AR,
DEFORMED,
190°C/15HRS



450°C/4HRS,
DEFORMED,
190°C/16HRS



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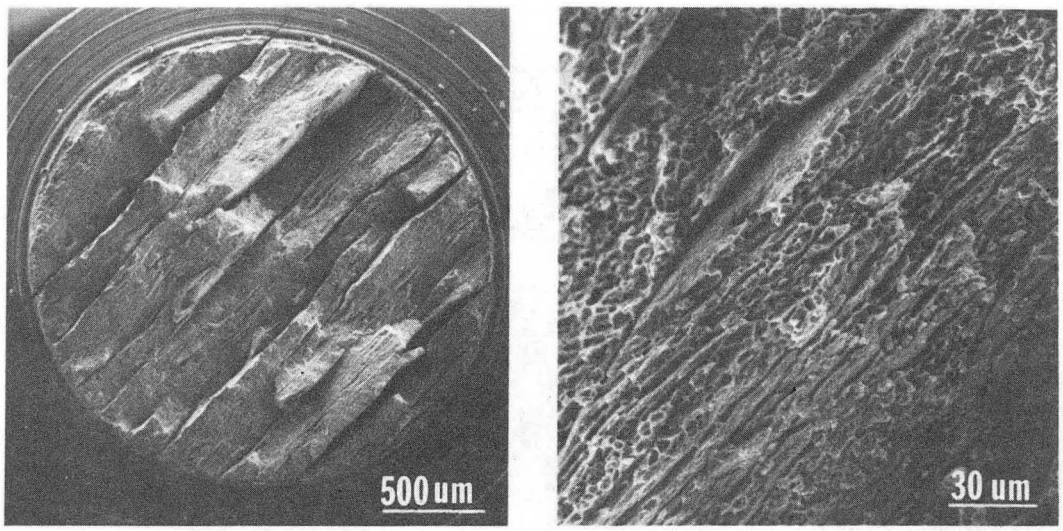
9. Notched tensile fracture modes of deformed alloys. (top) base alloy, (bottom) dispersoid alloy.

450°C/10HRS + 190°C/20HRS



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10. Notched tensile fracture mode of 450-A alloy.

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