LBL-25020 c.2

Center for Advanced Materials

CAM

Presented at the Materials Research Society International Meeting on Advanced Materials, Tokyo, Japan, May 30–June 3, 1988

RECEIVED
LAWRENCE
BERKELEY LABORATORY

OCT 11 1988

LIBRARY AND
DOCUMENTS SECTION

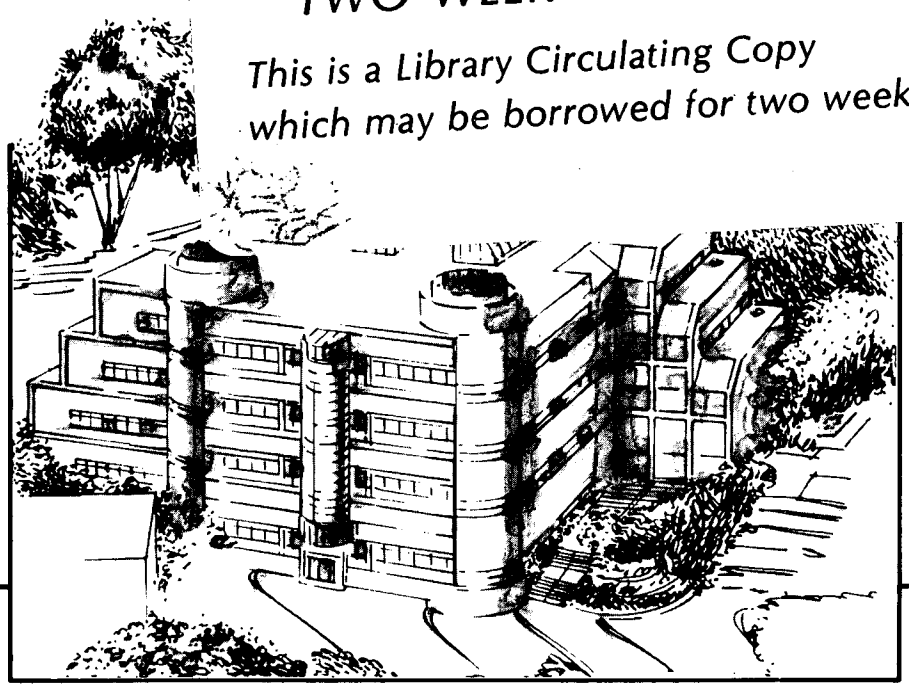
Mechanical Behavior of Superplastically-Formed Al-Cu-Li-Zr Alloy 2090

J. Glazer, T.G. Nieh, J. Wadsworth, and J.W. Morris, Jr.

May 1988

TWO-WEEK LOAN COPY

This is a Library Circulating Copy which may be borrowed for two weeks.



Materials and Chemical Sciences Division

Lawrence Berkeley Laboratory • University of California

ONE CYCLOTRON ROAD, BERKELEY, CA 94720 • (415) 486-4755

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

LBL-25020 c.2

DISCLAIMER

This document was prepared as an account of work sponsored by the United States Government. While this document is believed to contain correct information, neither the United States Government nor any agency thereof, nor the Regents of the University of California, nor any of their employees, makes any warranty, express or implied, or assumes any legal responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by its trade name, trademark, manufacturer, or otherwise, does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof, or the Regents of the University of California. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof or the Regents of the University of California.

LBL-25020

MECHANICAL BEHAVIOR OF SUPERPLASTICALLY-FORMED AL-CU-LI-ZR ALLOY 2090

J. Glazer,* T.G. Nieh,** J.Wadsworth** and J.W. Morris, Jr.*

* Center for Advanced Materials
Materials and Chemical Sciences Division
Lawrence Berkeley Laboratory
University of California

and

Department of Materials Science and Mineral Engineering
University of California at Berkeley
Berkeley, California 94720

** Lockheed Research and Development Division,
Metallurgy Department,
Palo Alto, California

May, 1988

This work was jointly supported by the Lockheed Independent Research Program and the Director, Office of Energy Research, Office of Basic Energy Science, Materials Sciences Division of the U.S. Department of Energy under Contract No. DE-AC03-76SF00098.
J. Glazer is an AT&T Bell Laboratories PhD Scholar.

Mechanical Behavior of Superplastically-Formed Al-Cu-Li-Zr Alloy 2090

J. Glazer,* T.G. Nieh,** J.Wadsworth** and J.W. Morris, Jr.*

* Center for Advanced Materials, Lawrence Berkeley Laboratory, and Department of Materials Science and Mineral Engineering, University of California at Berkeley

** Lockheed Research and Development Division, Metallurgy Department, Palo Alto, California

The design, manufacture and post-forming analyses of a complex shape produced by superplastic forming of the aluminum-lithium alloy 2090 are described. With appropriate prior thermomechanical processing to produce fine-grained superplastic sheet, this alloy displays excellent superplastic properties at commercially desirable strain rates and temperatures. The mechanical behavior at room and cryogenic temperatures of formed and unformed sheet after solution heat treatment and aging was investigated. In both cases, tensile properties improve significantly at low temperatures over their room temperature values. However, these increases are not accompanied by an increase in fracture toughness at low temperature over the moderate value measured at room temperature.

Introduction

Advanced aerospace launch systems of the future will incorporate large structural tanks containing liquid cryogenic fuels such as hydrogen (20K), oxygen (90K) or methane (110K). Cryogenic tankage is a demanding application that requires a material with good mechanical properties between room and cryogenic temperatures. Because of the premium on reducing structural weight of spacecraft, the specific properties must meet even more stringent criteria. Consequently, only low density alloy systems are feasible choices. In view of the increasing importance of lowering launch costs, ease of manufacture, cost and durability are also of great importance [1,2]. In this context, superplastic forming is an appealing manufacturing technology because it allows production of complex shapes with a minimum expenditure of material, energy and labor. Some of these parts may not be manufacturable by any other technique. Since various tank designs are still under consideration, the relative importance of the various criteria mentioned above is in a state of flux; however, it is possible to identify promising materials. One of these is the aluminum-lithium alloy AA 2090 (nominal composition Al-2.2Cu-2.7Li-0.1Zr in weight percent).

Previous work has demonstrated that this alloy has high specific strength, stiffness and toughness [3-6], excellent cryogenic mechanical properties [7-10] and, like other aluminum-lithium alloys, can be processed to display superplastic behavior [11,12]. This paper reviews a collaborative investigation designed to assess the commercial feasibility of producing formed parts for cryogenic tankage applications. Forming procedures required to produce a complex component from 2090 sheet, as well as the properties of the formed component at room and cryogenic temperatures, were investigated.

Experimental Procedure

The details of material preparation, the thermomechanical processing (TMP) of 2090 sheet to obtain superplasticity, and the actual forming process have been described in detail [13] and are only reviewed briefly here. The tensile behavior TMP 2090 at cryogenic temperature has also been reported previously [14]. The material was induction cast, homogenized 24 hours at 538°C and scalped after cooling to room temperature. The ingot was then hot rolled in two steps to 20 mm thickness. The chemical composition of the sheet was analyzed to be Al-2.6Cu-2.4Li-0.18Zr in weight percent.

Superplastic forming and subsequent processing

The TMP which led to optimum superplasticity in this alloy is a four-step (overage) procedure. The initial step is a solution heat treatment followed by a rapid quench to create a microstructure of a coarse-grained matrix containing ~0.1μm Al_3Zr particles. The second step, a severe overaging, leads to a uniform distribution of large diameter (≥ 0.5μm) equilibrium second phase particles. Warm rolling is carried out in the third step, to create intense deformation bands of high local defect densities around these large second phase particles. The alloy is recrystallized in the fourth step to complete the procedure. During recrystallization the large second phase particles dissolve. Growth of newly-recrystallized grains is inhibited by the pinning action of the fine Al_3Zr particles. The result of this procedure is the development of a very fine-grained structure. For AA 2090, the actual procedure was to solution heat treat at 538°C/1 hour, cold water quench, overage at 400°C/16 hours, air cool to 290°C and isothermally roll to 2 mm (true strain $\varepsilon = -2.3$) and, finally, recrystallize at the superplastic forming temperature, 485°C.

The excellent superplastic properties of the 2090 alloy in uniaxial tension [15] led to an evaluation of its formability under production conditions. Superform USA carried out the forming on specially processed sheet provided by Reynolds Metals Company. The sheet was superplastically formed at 485°C (905°F) into a headbox, a hat-shaped piece with a rectangular cross section. The part was formed by using a gas pressure to form the sheet into a bubble into which a male tool was inserted. The pressure was then gradually reversed to form the sheet around the top of the tool [13].

All material (formed or not) was solution heat treated 0.5 hours, quenched in 5% glycol in cold water, and aged to approximately peak strength at 190°C (375 F) for 8 hours and air cooled. The choice of aging time is discussed in a subsequent section. To examine the

evolution of the microstructure, optical microscopy was performed at various stages of the processing on specimens polished to 0.05μm and etched using Keller's reagent.

Mechanical testing

Tensile specimens were cut from the flange of the hat (superplastic forming true strain $\varepsilon = 0$) and from the sides and top of the hat which were 0.5 mm (0.020 in) and 0.75 mm (0.30 in) thick ($\varepsilon = -1.1$ and -0.9, respectively). Control specimens were cut from the unformed sheet which was approximately 2 mm (0.080 in) thick. The specimens were subsize flat tensile specimens with a 2.54 cm (1.0 in) gauge length and were cut parallel to the original rolling direction. Tensile tests were performed at room temperature, 77 K (liquid nitrogen) and 4 K (liquid helium) on a servo-hydraulic testing machine equipped for cryogenic testing. Elongations less than 20% were measured using an extensometer. All tests were conducted using a crosshead displacement speed of 2.1×10^{-3} cm/sec (0.05 in/min). All results are from duplicate tests except for the 75mm, 4 K datum.

Toughness was assessed using a Kahn-tear type specimen. This test has been used to determine the relative toughness of various aluminum alloys in the past [16,17]. The results of the test are reported as the tear strength, i.e. the load required to initiate crack growth from a sharp notch, and the unit propagation energy (UPE), which is the total energy required to propagate the crack. Because of the difficulty of obtaining samples of uniform processing from the headbox, only the unformed superplastic sheet was examined. Single specimens machined to 1.6 mm (0.063 in) thickness were taken from the sheet parallel to and perpendicular to the final rolling direction and tested at room temperature and 77 K.

Results

Formability

The specially-processed 2090 alloy showed excellent superplastic elongations in uniaxial tension of 900% at commercially viable strain rates and temperatures. The elongations compare well to those of SUPRAL (AA 2004, nominal composition Al-6Cu-0.4Zr) at similar strain rates (see figure 1). Because of their fine grain size, the usable strain rates of these alloys are an order of magnitude higher than for AA 7475 (Al-5.7Zn-2.2Mg-1.5Cu-0.2Cr), an important consideration for commercial forming. Trials conducted to establish the forming procedure indicated that the material could be formed readily under a range of forming parameters. This is an important development step since the stress states involved in forming the material into a complex shape are more complicated than those of the uniaxial tension tests typically used for laboratory applications.

Microstructure

A microstructural analysis of the formed material before and after heat treating was performed to assess the impact of the superplastic forming operation (see figure 2). Little difference was found between regions representing different degrees of thinning. In the as-

formed material, two second phases, both uniformly distributed in the matrix, are distinct in optical micrographs. In addition, large inclusions are also present; energy dispersive x-ray analysis (EDX) suggested they are of the Al_7Cu_2Fe type. Of the matrix phases, it was found that the discontinuous phase was richer in Cu than the continuous phase. Analysis for Li was not undertaken. The microstructure was significantly altered by the T6 heat treatment. Numerous rod-like precipitates occur on the grain boundaries; these phases are presumed to be equilibrium T_1 and/or T_2 phases. The Al_7Cu_2Fe particles appear unchanged by the heat treatment.

Qualitative observations of the degree of cavitation in the as-formed material were also made (see figure 3). In regions of moderate superplastic strain (e.g. $\varepsilon = -0.9$) cavitation is minimal although localized regions of cavities are observed. However, at the highest strains ($\varepsilon = -1.1$) there was significantly more cavitation. This area is located at the intersection with the flange, and thins excessively due to friction with the tool. The majority of cavities occur at the centerline.

Aging

An aging study was conducted to determine the T6 condition for this alloy after superplastic forming under the condition described above. To investigate the influence of varying amounts of superplastic deformation, samples from four locations on the headbox were compared with the behavior of the as-received material. Aging was conducted at 160, 171 and 190°C. Aging response was measured using a Rockwell superficial hardness tester after lightly grinding the surface to remove any oxide layer. There was little difference between the as-received sheet and the flange (which underwent the forming cycle, but was not deformed). However, for all heat treatment times and temperatures, a consistent decrease in strength with decreasing thickness in the as-formed condition was noted. There was also a variation in the rate at which aging occurred with amount of deformation. At 160°C, overaging had not occurred after 128 hours. At 171°C, some thicknesses appear to peak at 64 hours while others are not overaged at 128 hrs. At 190°C the peak occurs between 8 and 16 hours and the material softened rapidly with further aging. The maximum attainable strength was not a strong function of aging temperature. The optimum heat treatment was selected to maximize the strength of the thinnest material, since the load at which it yields could well be limiting in service, and minimize the heat treatment time (to minimize manufacturing costs). An aging treatment of 190°C/8 hours was chosen.

Mechanical Properties

The tensile properties of the material are given in Table I and plotted in figure 4. Properties for 2219-T51 [18] and 2090-T81 [9] are included for comparison. The material showed decreasing yield and ultimate tensile strengths with decreasing thickness (and increasing amount of superplastic deformation) at all temperatures. This decrease is at least partially attributable to increased cavitation in more severely formed regions which may effectively reduce the load-bearing cross section of the specimen by ten or more percent of its nominal value. The yield strengths increase about 15% between room temperature and 4 K, while the ultimate tensile strength increases somewhat more rapidly. The improvement in elon-

gation between room temperature and 4 K is particularly striking and is a factor of six in some cases. The uniform elongation varies considerably between specimens, but does not appear to be greatly affected by thickness. The majority of the specimens failed without necking. However, analysis of the strain hardening behavior using the procedure described in ref. 10 indicated that the specimens failed at the tensile instability point. The increased elongation at low temperature seemed to be associated with increased strain hardening rates at low temperature that postponed tensile instability.

Table I. Tensile properties of 2090 sheet before and after superplastic forming. Strain ε is true strain during superplastic forming. Data for 2090-T81 and 2219-T851 (plate, longitudinal orientation) are taken from refs. 9 and 18, respectively.

Specimen	ε	Test Temp. (K)	σ_y MPa (ksi)	σ_{uts}¥ MPa (ksi)	uniform elong. %
Sheet	0	298	315 (46)	415 (60)	4.2
		77	350 (51)	510 (74)	11.3
		4	415 (60)	650 (94)	25
Flange	0	298	310 (45)	405 (58.5)	4.4
		77	345 (50)	510 (74)	13.4
		4	370 (54)	600 (87)	17
0.75 mm	-1.0	298	285 (41)	380 (55)	5.1
		77	315 (46)	480 (70)	10.5
		4	370 (54)	450 (65)	2
0.50 mm	-1.4	298	260 (38)	360 (52)	4.8
		77	275 (40)	>415 (60)†	>5.6†
		4	240 (35)*	490 (71)*	19*
2090-T81		298	455 (66)	490 (71)	6
		77	500 (73)	625 (91)	14
		4	535 (78)	650 (95)	20
2219-T851		298	371 (54)	466 (68)	11
		77	440 (63)	568 (82)	14
		4	484 (70)	659 (96)	15

¥ Engineering stress computed from highest measured load.
† Specimen thickness along gauge length was uneven. Elongation was non-uniform.
* Specimen was taken from a second lot of material of lower strength.

Table II. Results of Kahn-type tear tests on unformed superplastic sheet aged to -T6 condition.

Test temperature (K)		Tear Orientation	Tear/yield strength ratio (MPa)	Unit propagation energy (kg-m/m^2)
298	L-T	335	1.05	95
	T-L	310	1.00	105
77	L-T	350	1.00	--*
	T-L	360	1.00	45

* Invalid test.

The results of the Kahn tear tests are given in Table II. Because the recrystallized microstructure is relatively equiaxed, the L-T and T-L properties should be roughly equivalent. Kaufman and Holt [16] have established a correlation between K_c and UPE. Although the limited data set precludes firm conclusions, the K_c apparently decreases, although not catastrophically. It is difficult to make numerical comparisons because the UPE is higher if the test is performed using an elastically hard testing machine (used by Kaufman and Holt) than the apparent UPE measured using an elastically soft machine (used in this work). Applying the correlation in ref. 16 to these data therefore underestimates the K_c, perhaps considerably. With these caveats, the K_c values are estimated to be > 80 MPa$\sqrt{m}$ at room temperature and > 60 MPa$\sqrt{m}$ at 77 K.

Fracture Behavior

Any variation in failure mode should also provide some insight into the variation of mechanical properties with temperature. The most obvious difference is that all tensile specimens tested at room temperature failed in shear at 45° to the tensile axis, whereas specimens broken at 4 K exhibited flat intergranular fracture surfaces. The failure at liquid nitrogen appears to be a mixture of the two failure modes. The tensile fracture appearance of the unformed superplastic sheet is illustrated in figure 5. The fracture appearance of the Kahn tear specimens also exhibited a transition with temperature, in this case from ductile dimple rupture at room temperature to a mixed mode at 77 K which combined small, shallow dimples with small regions of intergranular failure. This transition, which was accompanied by a roughening of the fracture surface at 77 K that suggested that the crack was seeking weak points in the microstructure.

Discussion

The mechanical properties of superplastically formed 2090-T6 would appear to be promising for commercial applications both at room and cryogenic temperatures. Although the room temperature yield and tensile strengths are significantly depressed below 2090-T81, they are superior to those of SUPRAL, which is currently used to produce superplastically

formed parts. Furthermore, the material may compare better with competing materials at cryogenic temperature than at room temperature. For example, in comparison with 2219-T87, a non-superplastic alloy used in many cryogenic applications, the room temperature strengths are low; however, at 4 K the ultimate tensile strength is comparable to that of 2219 (and the uniform elongation considerably greater). Further study of the influence of more complex post-forming aging treatments would probably result in improved yield strengths.

Summary

1. Appropriately processed 2090 can be superplastically formed using commercial practice to make complex parts.
2. The tensile yield strength, ultimate strength and uniform elongation all increase with decreasing temperature. A slight drop in strength is observed in severely formed regions.
3. Moderate toughness at room temperature is indicated. The toughness appears to decline at lower temperature, but not catastrophically.

Acknowledgement

The authors would like to thank Ms. C. Henshall and Ms. C. McNally, Research and Development Division, Lockheed Missiles and Space Co., Inc., and Ms. S.L. Verzasconi for assistance during the course of this study. The authors would also like to thank Mr. Bennie Ward and his colleagues at Reynolds Metals Co. for supplying the 2090 sheet. This study was jointly funded by the Lockheed Independent Research Program and the Director, Office of Energy Research, Office of Basic Energy Science, Material Sciences Division of the U.S. Department of Energy under Contract No. DE-AC03-76SF00098. J. Glazer is supported by an AT&T Bell Laboratories PhD Scholarship.

References

1. A. Taylor, ASM WESTEC Aluminum-Lithium Symposium, 1988.

2. See for example, C. Covault, Aviation Week and Space Tech., April 27, 1987, p. 32.

3. "Aluminum-Lithium Alloys I," T.H. Sanders, Jr. and E.A. Starke, Jr., eds.,TMS-AIME, Warrendale, PA (1982)

4. "Aluminum-Lithium Alloys II," T.H. Sanders, Jr. and E.A. Starke, Jr., eds.,TMS-AIME, Warrendale, PA (1984)

5. "Aluminium-Lithium Alloys III," C. Baker, P.J. Gregson, S.J. Harris and C.J. Peel, eds., The Institute of Metals, London (1986).

6. "4th International Aluminium Lithium Conference," G. Champier, B. Dubost, D. Miannay and L. Sabetay, eds., Journal de Physique, Colloque C3, supplément au n°9, Tome 48, September 1987.

7. J. Glazer, S.L. Verzasconi, E.N.C. Dalder, W. Yu, R.A. Emigh, R.O. Ritchie and J.W. Morris, Jr., Adv. Cryo. Eng., 32, 397 (1986).

8. R.C. Dorward, Scripta Met., 20, 1397 (1986).

9. J. Glazer, S.L. Verzasconi, R.R. Sawtell, and J.W. Morris, Jr., Metall. Trans. 18A 1695 (1987).

10. K.T. Rao, H.F. Hayashigatani, W. Yu and R.O. Ritchie, Scripta Metall., 22 93 (1988).

11. J. Wadsworth, C.A. Henshall and T.G. Nieh, "Aluminium-Lithium Alloys III," C. Baker, P.J. Gregson, S.J. Harris and C.J. Peel, eds., The Institute of Metals, London, p. 199 (1986).

12. J. Wadsworth, A.R. Pelton and R.E. Lewis, Met. Trans, 16A, 2319 (1985).

13. C.A. Henshall, J. Wadsworth, M.J. Reynolds and A.J. Barnes, Materials and Design, 8, 324 (1987).

14. J. Glazer, T.G. Nieh and J.W. Morris, Jr., Adv. Cryo. Eng. 34, 291 (1987).

15. J. Wadsworth, C.A. Henshall, A.R. Pelton and B.Ward, J. Mater. Sci. Lett., 4, 674 (1985).

16. J.G. Kaufman and M. Holt, "Fracture Characteristics of Aluminum Alloys," Alcoa Research Laboratories Technical Paper No. 18, Aluminum Company of America, Pittsburgh, PA (1965).

17. J.G. Kaufman and M. Holt, Adv. Cryo. Eng. 10, 77 (1965).

18. J.G. Kaufman, K.O. Bogardus, and E.T. Wanderer, Adv. Cryo. Eng. 13, 294 (1968).

Figure Captions

Figure 1. Superplastic elongations as a function of strain rate for various aluminum alloys.

Figure 2. Optical micrograph of a region superplastically formed to $\varepsilon = -0.8$ showing second phase particles present in T6 condition.

Figure 3. Optical micrographs illustrating cavitation developed during forming. a) true superplastic strain $\varepsilon = -1.1$, b) $\varepsilon = -0.9$

Figure 4. Tensile properties of superplastically formed 2090-T6 as a function of temperature.

Figure 5. Scanning electron micrographs showing fracture appearance transition from shear to intergranular fracture in unformed tensile specimens broken at room temperature and 4 K.

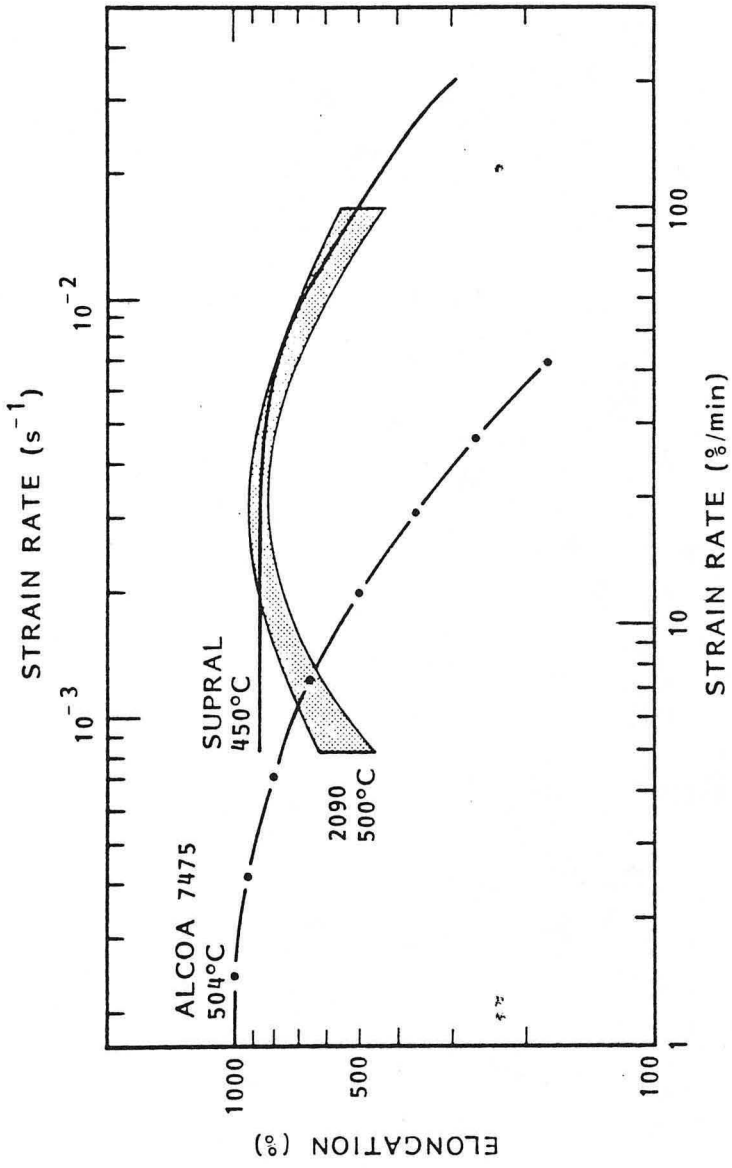
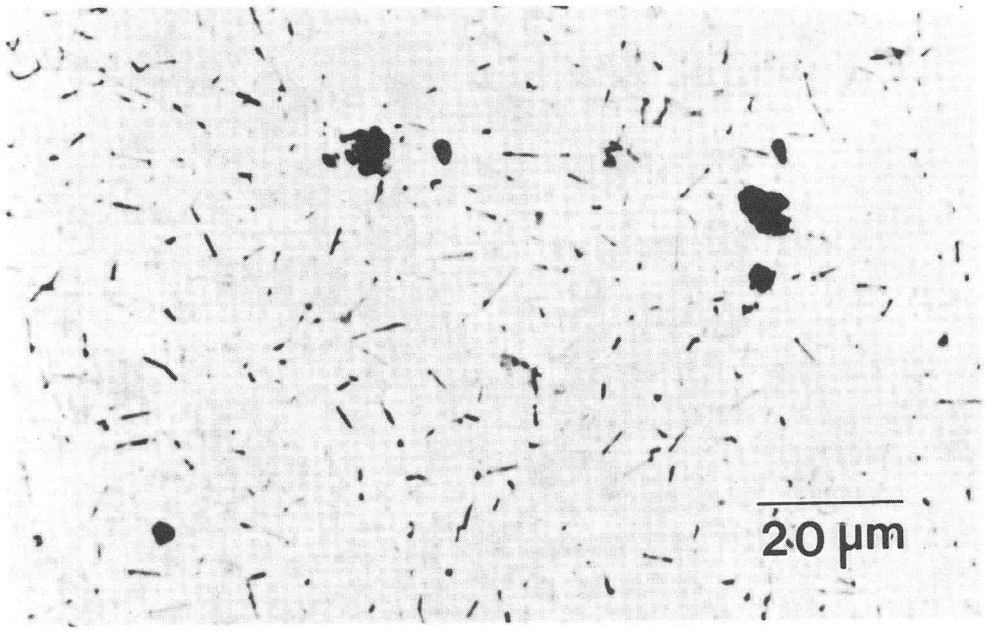
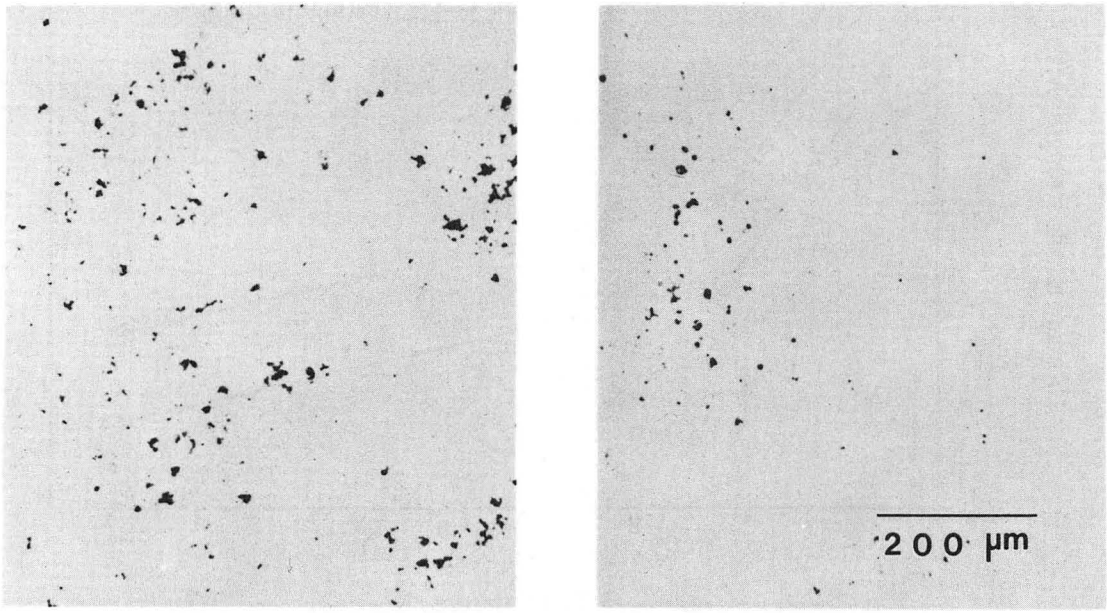

STRAIN RATE (s^{-1})
10^{-3}
10^{-2}
ALCOA 7475
504°C
SUPRAL
450°C
2090
500°C
1000
500
100
ELONGATION (%)
1
10
100
STRAIN RATE (%/min)

Figure 1



XBB 886-6205

Figure 2



XBB 885-4847

Figure 3

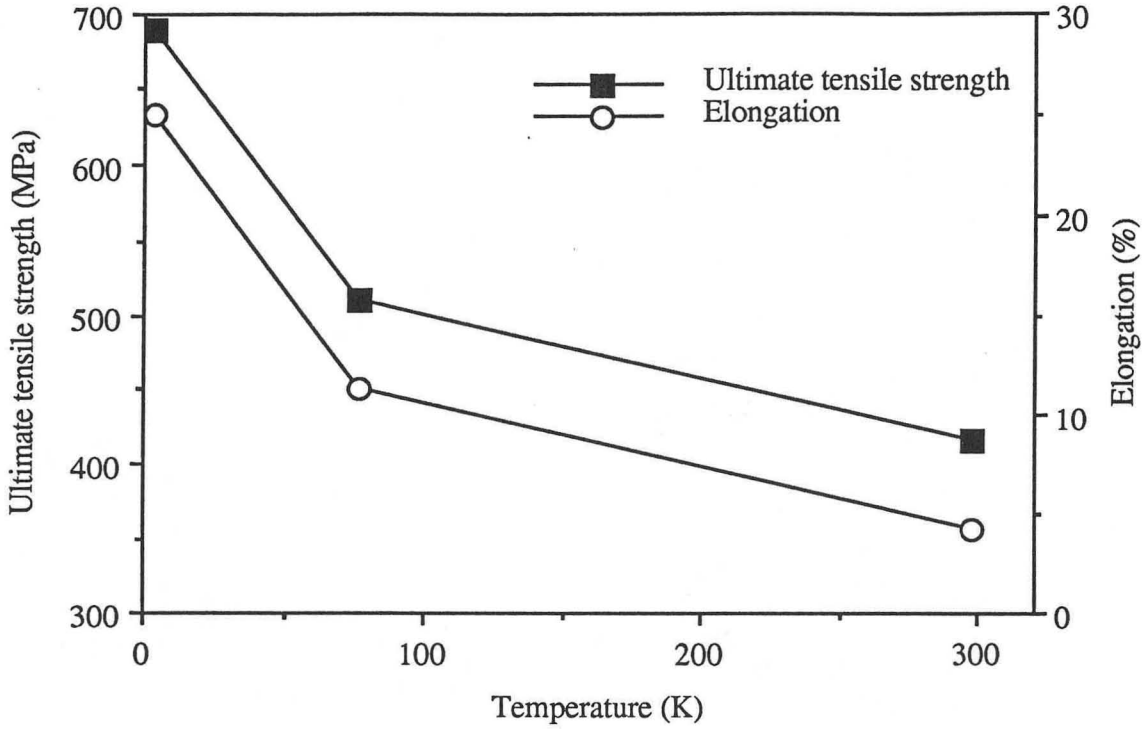
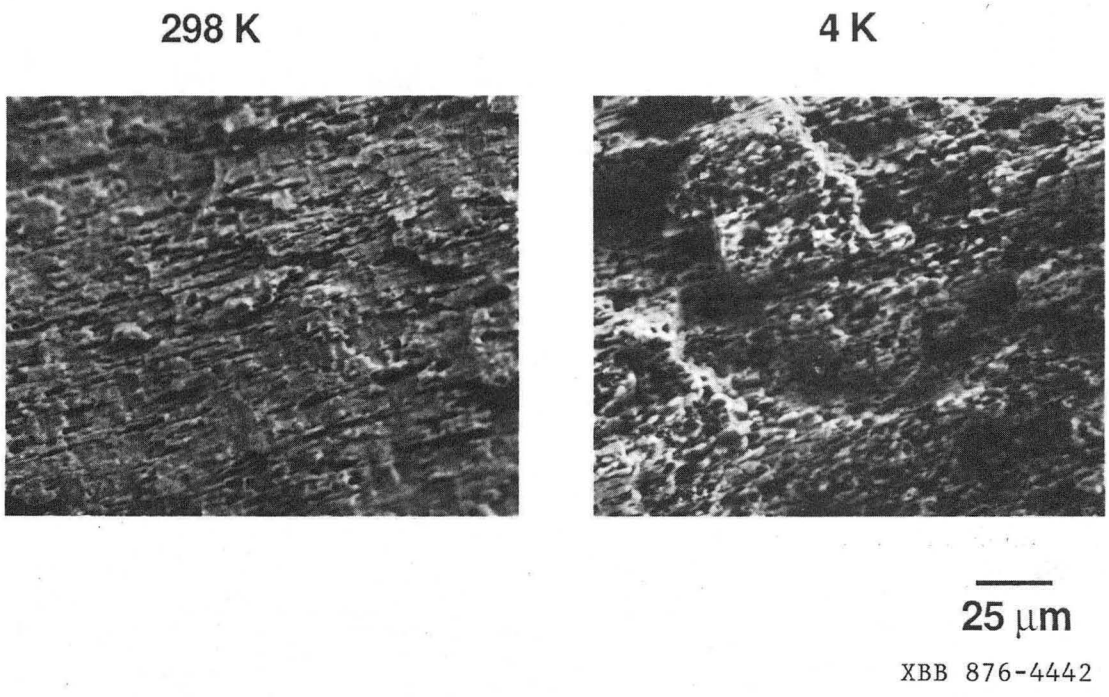

700
600
500
400
300
30
20
10
0
0
100
200
300
Ultimate tensile strength (MPa)
Elongation (%)
Temperature (K)
Ultimate tensile strength
Elongation

Figure 4



XBB 876-4442

Figure 5

LAWRENCE BERKELEY LABORATORY
TECHNICAL INFORMATION DEPARTMENT
UNIVERSITY OF CALIFORNIA
BERKELEY, CALIFORNIA 94720