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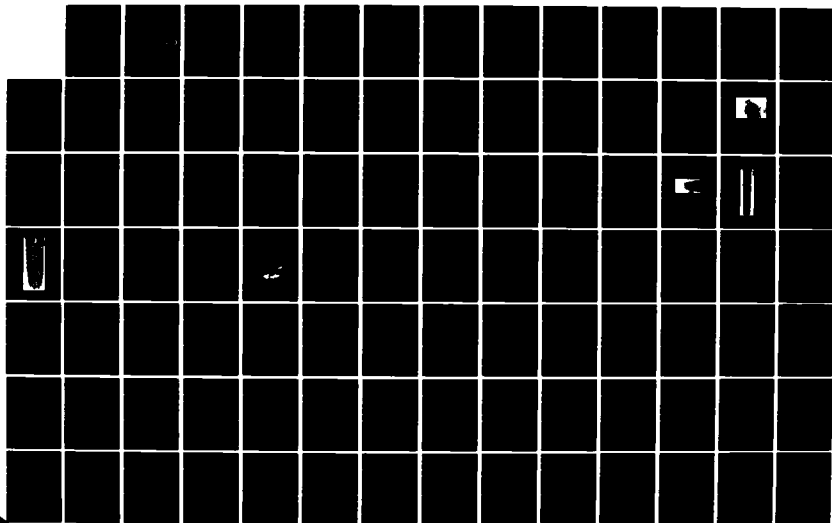
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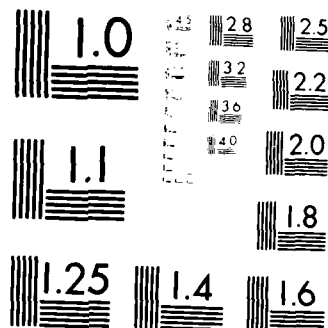
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SUPERPLASTICITY IN A THERMO-MECHANICALLY
PROCESSED ALUMINUM-10.2%Mg-0.52%Mn ALLOY

by

Max E. Mills

September 1984

Thesis Advisor:

T. R. McNelley

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20. ABSTRACT (continued)

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Superplasticity in a Thermo-Mechanically Processed
Aluminum-10.2%Mg-0.52%Mn Alloy

by

Max E. Mills
Lieutenant Commander, United States Navy
B.S.CH.E., University of Idaho, 1976

Submitted in partial fulfillment of the
requirements for the degree of

MASTER OF SCIENCE IN MECHANICAL ENGINEERING

from the

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September 1984

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This research extended the previous work performed by Becker on the elevated temperature deformation characteristics of an aluminum-10.2% magnesium-0.52% manganese alloy. The alloy was warm rolled at 300 C to 94% reduction. Stress-strain testing was utilized to collect data for stress vs strain rate and ductility vs strain rate, as well as, stress exponents and activation energies. Tensile testing was performed at strain rates from $1.39 \times 10^{-4} \text{ s}^{-1}$ to $1.39 \times 10^{-2} \text{ s}^{-1}$ and temperatures from 20 C to 425 C. Ductility ranged from 400% at 300 C and 600% at 325 C to 700% at 425 C. The data clearly establishes that the warm rolled alloy is superplastic at temperatures as low as 275 C and may exhibit superplastic elongations (greater than 400%) at strain rates as high as 10^{-2} s^{-1} at 325 C. Scanning electron microscope observations indicated little or no void formation at or below 300 C. The high ductilities observed at temperatures above the solvus are the result of grain boundary sliding.

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1. INTRODUCTION

The purpose of this thesis was to investigate the elevated temperature deformation characteristics of a thermo-mechanically processed Al-10.2%Mg-0.52%Mn alloy in the as-rolled condition. Previous research by Ness [Ref. 1], Bingay [Ref. 2], Glover [Ref. 3], Grandon [Ref. 4], Speed [Ref. 5], Chesterman [Ref. 6], Johnson [Ref. 7], and Shirah [Ref. 8], clearly demonstrated that thermo-mechanically processed aluminum-magnesium alloys exhibit high strength with good ductility at ambient temperature. Transmission electron microscopy studies by McNelley and Garg [Ref. 9] confirmed that the microstructure of this alloy consisted of fine, cellular dislocation structures or subgrain structures. It was further observed that annealing after rolling resulted in recovery with possible recrystallization to fine grains of submicron size. This prompted study of the elevated temperature behavior of these alloys with a view toward their possible superplastic behavior.

Although his findings were preliminary, Becker [Ref. 10] observed superplasticity in both the 8% and 10% aluminum-magnesium alloys. These alloys were thermo-mechanically processed by warm rolling. Testing was conducted on material in the as-rolled condition, after

annealing for various times at 300 C, and in a recrystallized condition obtained by heating for one half hour at 440 C.

The processing technique developed by Johnson [Ref. 7] and the tensile testing procedure developed by Becker [Ref. 10] were used in the study of this 10.2%Mg-0.52%Mn aluminum alloy. Tensile testing was conducted using an electro-mechanical Instron machine with a Marshall three zone clamshell furnace for temperature control. The microstructure of the elongated test samples was examined using optical microscopy.

This thesis presents data obtained from the mechanical testing of an as-rolled magnesium-aluminum alloy as well as results of microstructural examination to assist in evaluation of mechanical test results.

II. BACKGROUND

A. ALUMINUM-MAGNESIUM ALLOYS

Aluminum alloys have been extensively studied because of their low density, ductility, and toughness. The higher strength alloys derive their strength mainly through precipitation and solid solution hardening. The formation of a second phase retards dislocation motion and grain growth.

Aluminum-magnesium alloys are characterized by a good strength to weight ratio, superior ductility, lower density with better corrosion resistance than other, higher strength aluminum alloys, and also good high cycle fatigue behavior. The strength can be improved through cold working.

B. PREVIOUS WORK

Ness [Ref. 1] initiated the investigation of high magnesium alloys at this laboratory. He studied an 18% aluminum-magnesium alloy, attempting to parallel the concepts developed by Bly, Sherby, and Young [Ref. 11] in their work on high carbon steel. Through mechanical working of an Fe-C material in the two phase ferrite-carbide region they obtained microstructural refinement and improved mechanical properties, as did Ness [Ref. 1] with a resulting compression

strength of 655 Mpa (99 KSI) in this very high magnesium alloy.

Bingay [Ref. 2] and Glover [Ref. 3] attempted variations in thermo-mechanical processing of aluminum alloys to eliminate cracking during the rolling process. To refine the microstructure, Bingay [Ref. 2] performed both isothermal and non-isothermal forging prior to rolling in 15-19% magnesium containing alloys. Processing was difficult and subsequent work shifted to examine lower magnesium alloys. Aluminum alloys containing 7-9% magnesium were investigated by Glover [Ref. 3] who observed the characteristics of superplastic behavior.

Extending the study into 7-10% magnesium alloys, Grandon [Ref. 4] introduced a 24 hour solution treatment followed by a quench and then warm rolling at 300 C. His results indicated a doubling of strength while maintaining good ductility when compared to the 5xxx series. He also noted that recrystallization did not occur during warm rolling below the solvus. Alloys with greater magnesium content were tested by Speed [Ref. 5].

Chesterman [Ref. 6] studied the nature of precipitation and recrystallization in these alloys through optical microscopy. For 8-14% alloys, he reported that recrystallization only occurred at temperatures above the solvus and was apparently not induced even after extensive cold working followed by annealing as long as the annealing temperature

was below the solvus. At annealing temperatures of $.6T_m$, precipitation still replaced recrystallization as the method of stored energy release.

Johnson [Ref. 7] standardized the thermo-mechanical processing of the aluminum-magnesium alloys. He investigated 8-10% alloys and reported material strength of twice that of the 5xxx alloys with good ductility. His process was to solution treat the material at 440 C for 9 hours, isothermally upset forge the material, anneal for 1 hour at 440 C, quench, and then warm roll. Various warm rolling temperatures were used ranging from 200 C to 340 C. He concluded that the beta phase contributed by dispersion strengthening to the high strength and good ductility.

Shirah [Ref. 8] improved the microstructural homogeneity by increasing the solution treatment time to 24 hours. This extended treatment minimized precipitate banding while not effecting grain growth.

Becker [Ref. 10] combined the previous studies and developed the procedure for tensile testing isothermally at various temperatures up to 300 C. His work concentrated on 8.14% Mg and 10.2% Mg aluminum-magnesium alloys. He observed superplastic elongations of up to 400% and concluded that the higher magnesium content in the 10.2% alloy stabilized grain size and extended the range of superplastic behavior to higher temperatures.

C. SUPERPLASTIC BEHAVIOR

Superplasticity is the capability of a material to deform to exceptionally high elongations. Elongation greater than 200% is often taken as superplastic [Ref. 11]; values of greater than 1000% have frequently been reported. The most important prerequisites for superplasticity are generally agreed to be fine, equiaxed grains with high angle grain boundaries, temperature in the range of $0.5-0.7T_m$, low strain rates, and a high strain rate sensitivity coefficient (m).

To achieve superplasticity, a fine grain size of less than 10 microns is normally required. A second phase at the matrix grain boundaries is usually necessary to retard grain growth under warm temperature conditions. This second phase should be similar in strength to the matrix to minimize the formation of cavities [Ref. 11]. The fine grains should also be equiaxed with smooth, rounded grain boundaries to promote sliding. Grain growth suppresses superplasticity as larger grains impose larger diffusion distances and reduce the strain resulting from boundary sliding.

Deformation at elevated temperature is a thermally activated process, and superplasticity is observed only at elevated temperatures. The flow stress is a function of strain, strain rate, and temperature. At constant strain

and temperature, stress is often assumed to depend upon strain rate according to

$$\dot{\epsilon} = K \dot{\epsilon}^m \quad (\text{eqn 2.1})$$

where σ is the stress, $\dot{\epsilon}$ is the strain rate, K is a temperature dependent constant, and m is the strain rate sensitivity coefficient.

In general, m increases with increasing temperature. Superplastic behavior in metals usually occurs at high m values of .3 to .5 and is the greatest at the maximum m . The value of m can be found by plotting log stress vs log strain rate for data obtained at constant strain and temperature.

The activation energy (Q) is a measure of the energy required for temperature-dependent processes. For a thermally activated deformation process

$$\dot{\epsilon} = f(\sigma) \exp(-Q/RT) \quad (\text{eqn 2.2})$$

where R is a gas constant and T is temperature.

Values for the activation energy can be obtained by plotting log strain rate vs inverse temperature for data at constant stress. Such plots often indicate that the activation energy may be constant for a range of stress but may change to a different value for a different range of stress. Values of activation energy for deformation often are the same as those for lattice diffusion, suggesting

lattice diffusion control of deformation. When grain boundary sliding controls the deformation, lower values of activation energy may be observed; diffusion in the grain boundaries, the rate controlling process, occurs more readily than diffusion in the grain interior and is characterized by the lower activation energy. Hence activation energy measurement may provide useful insight into the mechanism involved in the material.

III. EXPERIMENTAL PROCEDURE

A. MATERIAL PROCESSING

The aluminum alloy investigated was nominally 10.2 weight % Mg and 0.52 weight % Mn. The direct chill cast ingot received was produced by ALCOA Technical Center using 99.99% pure aluminum base metal alloyed with commercially pure magnesium, 5% beryllium-aluminum master alloy [Ref. 7]. The ingot measured 127 mm (5 in) in diameter and 1016 mm (40 in) in length. The complete composition is listed below in Table I [Ref. 7].

Table I
Alloy Composition (Weight Percent)

<u>Serial Number</u>	<u>Si</u>	<u>Fe</u>	<u>Mn</u>	<u>Mg</u>	<u>Ti</u>	<u>Be</u>
501300A	0.01	0.03	0.52	10.2	0.01	0.0002

A portion of the ingot was sectioned into billets 101.6 mm (4 in) long with a square cross section of width 31.75 mm (1.25 in). Following the procedures developed by Johnson [Ref. 7] and Becker [Ref. 10] the billets were solution treated at 440C for 24 hours, upset forged at 440C on heated platens to approximately 28 mm (1.1 in) in height, annealed at 440C for 1 hour, then oil quenched. The billets were

forged along their greatest dimension, resulting in a reduction of approximately 73% or a strain of about 1.3.

B. WARM ROLLING

The rolling of the billets into sheets was comparable to that done by Becker [Ref. 10] and performed in the same manner as Johnson [Ref. 7], with some modifications made to the technique. To preclude cracking of the forged billets by suspected uneven heating, each billet was placed on a steel plate used as a heat source in the rolling furnace. Since isothermal heating was essential each sample was initially heated from room temperature to 300C for approximately 10 minutes (after the sample surface temperature reached 300C) before attempting the first rolling pass. The samples were then heated for 6 minutes between passes for the first three passes and 1 to 3 minutes between passes thereafter, leaving the sample in the furnace just long enough to ensure an isothermal condition. The temperatures of both the sample and the plate were monitored using thermocouples. In the later rolling passes, the deformed sheet was pulled through the rollers to minimize warping. Each billet was rolled into a sheet about 1.8 mm (0.07 in) thick, 102 mm (4 in) wide, and 762 mm (40 in) long resulting in a final sample reduction of approximately 94%.

As described in Becker [Ref. 10], the rolled sheets were cut into blanks 63 mm (2.47 in) long and 13 mm (0.5 in)

wide. Depending upon sheet thickness each sheet yielded 130 to 140 blanks. The blanks were endmilled in lots of five to a gage width of approximately 3 mm (0.12 in) and length of 15 mm (0.6 in). A fabricated jig was used as a milling guide in determining the gage dimensions. A sketch of the test specimen is shown in Figure 3.1.

C. SPECIMEN TESTING

The tensile testing procedure was similar to that described by Becker [Ref. 10]. Each test specimen was placed into wedge grips and held in place by pins passing through the wedges. The wedge-action grips, grip assemblies, and pull rods were supplied by ATS, King of Prussia, PA, and were fabricated from Inconel 718. The assembly was then mounted into pull rods connected to an electro-mechanical Instron machine.

To maintain a constant specimen temperature during testing, a Marshall clamshell furnace containing three, vertically oriented heating elements was utilized. The heating elements were individually regulated by three controllers using thermocouples located adjacent to each element. The furnace was insulated by positioning insulation paper between both halves of the clamshell and placing crescent-shaped insulation inside the top and bottom of the furnace. Rings of insulation were wrapped around each pull rod just outside the furnace, with a final

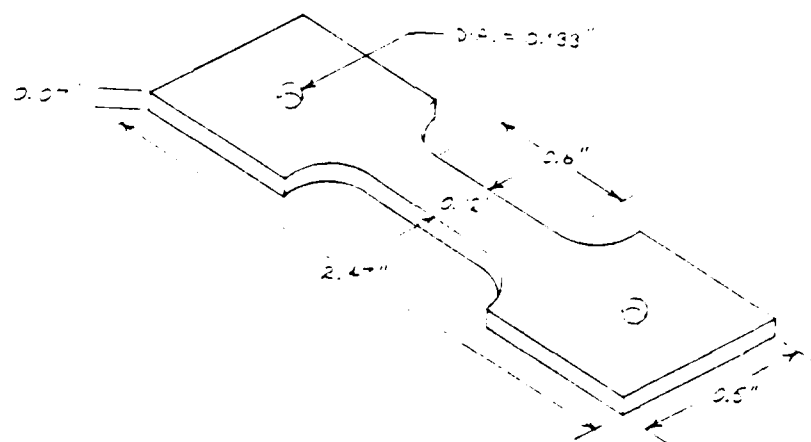


Figure 3.1 Test Specimen Geometry

insulation pad wrapped around the rings. The bottom insulation was wired to the furnace to keep it in place during testing.

Five thermocouples were placed inside the furnace to monitor the specimen temperature. A thermocouple was attached to each pull rod approximately four inches above and below the specimen and two additional thermocouples were touching the top and bottom of the specimen, respectively. A center thermocouple was initially placed touching the middle of the specimen but this tended to bend the sample and result in premature fracture. The thermocouple was subsequently positioned beside the specimen at the start of each test.

The tensile testing was performed with crosshead speeds ranging from 0.127 mm per minute to 127.0 mm per minute (0.005 in/min to 5.0 in/min) at temperatures from 20C to 425C. Care was taken to ensure each specimen was pulled isothermally. The testing apparatus was heated for a full day prior to conducting a sequence of experiments at a constant temperature. A test specimen was then mounted into the pull rods and the furnace closed. The five thermocouples indicated temperature equalization in approximately 1 hour and the test was started. At very low strain rates the bottom pull rod temperature would slowly start to drop as the bottom rod came out of the furnace. The top four temperatures normally remained identical, with the bottom

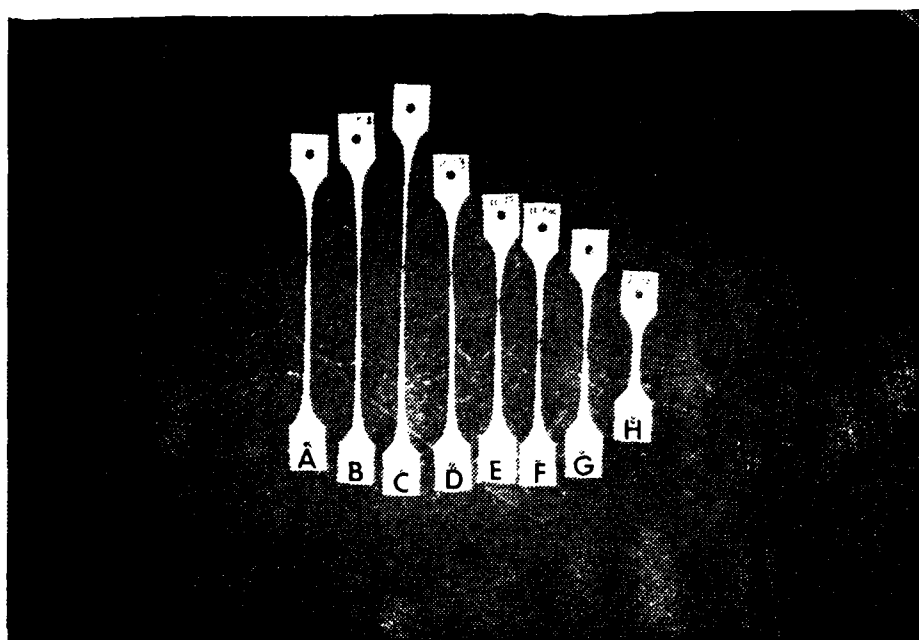
pull rod temperature dropping by no more than 10C before completing the test. Figure 3.2 is a visual summary of one test sequence.

D. DATA REDUCTION

The Instron strip chart recorder registered applied force as a function of chart motion. The magnification ratio between chart speed and crosshead speed varied but was usually set at 10. From the raw data, engineering stress and strain were computed and loaded into computer data files for plotting and further calculations. To remove such variables as grip tightening, Instron machine error, and elastic strain, a "floating slope" calculation was made at each selected data point using a computer subroutine. Sample elongation was found by measuring the fractured specimen.

E. COMPUTER PROGRAMS

All plotting and true stress-true strain calculations were accomplished using FORTRAN computer programs in conjunction with the library routine DISSPLA. Essentially, the appropriate input data files were read into each program and loaded into arrays. These arrays were then operated on to achieve the desired variables, loaded into DISSPLA, and plotted against each other. Also, various DISSPLA curve fitting routines were used on some of the plots to obtain



Key

- A. $1.4 \times 10^{-4} s^{-1}$
- B. $5.6 \times 10^{-4} s^{-1}$
- C. $1.4 \times 10^{-3} s^{-1}$
- D. $5.6 \times 10^{-3} s^{-1}$
- E. $1.4 \times 10^{-2} s^{-1}$
- F. $5.6 \times 10^{-2} s^{-1}$
- G. $1.4 \times 10^{-1} s^{-1}$
- H. Untested Sample

Figure 3.2 Photograph of Samples Tested at 325 C

smooth curves between data points. The computer programs are listed in Appendix B.

F. METALLOGRAPHY

Selected specimens were cold mounted on a base of glass using steel blanks or brass rings as a mold, depending upon the size of the sample. The mounted samples were then ground using 240 to 600 grit paper and polished with a magnesium oxide abrasive system. A Graf-Sargent solution was used to etch each specimen. The technique used was to swab each sample for 40 seconds. Using a Zeiss Universal Microscope, optical micrographs were taken with Panatomic X 35 mm film.

IV. RESULTS AND DISCUSSION

A. MECHANICAL TESTING RESULTS

To study the deformation characteristics of this alloy, tensile testing was conducted over a wide range of temperatures and strain rates using the procedures described in Chapter III. Temperatures varied from 20 C to 425 C and strain rates from 1.4×10^{-4} to $1.4 \times 10^{-1} \text{ s}^{-1}$ as illustrated in Table II.

True stress and true plastic strain were computed as described in Chapter III and plotted for each test temperature. One example is shown in Figure 4.1 for testing at 300 C, and the remainder of the data obtained is given in Appendix A. The curves drawn reflect data points taken prior to the onset of necking; this procedure was necessary as the assumption of uniform straining of the gage section does not apply once necking has begun. As often noted in studies of superplastic materials, the test samples exhibit prolonged necking during deformation. Particular attention was directed to the temperature interval from 200 C to 325 C since Becker's [Ref. 10] data indicated superplastic behavior in this region.

In this temperature range, the stress-strain plots for all temperatures indicate that at high strain rates a strain softening occurs as stress decreases significantly with

Table II

Data for Al-10.2%Mg-0.52%Mn Alloy in the As-Rolled Condition

Temperature C ⁻¹ Strain Rate s ⁻¹		UTS Mpa	True Stress at 0.1 Plastic Strain Mpa	Ductility
20	5.6X10 ⁻⁴	414.0	*	3.0
	5.6X10 ⁻³	478.0	*	3.2
	5.6X10 ⁻²	503.0	*	3.2
100	5.6X10 ⁻⁴	404.0	484.0	34.2
	5.6X10 ⁻³	453.0	524.0	22.5
	5.6X10 ⁻²	486.0	*	9.3
150	5.6X10 ⁻²	247.0	297.0	67.0
	5.6X10 ⁻³	327.0	386.0	51.5
	5.6X10 ⁻⁴	376.0	438.0	37.7
	1.4X10 ⁻¹	405.0	*	28.2
200	1.4X10 ⁻⁴	80.6	96.0	134.0
	5.6X10 ⁻⁴	119.0	150.0	187.0
	1.4X10 ⁻³	146.0	184.0	125.0
	5.6X10 ⁻³	166.0	209.0	144.0
	1.4X10 ⁻²	225.0	266.0	94.0
	5.6X10 ⁻²	268.0	297.0	58.8
	1.4X10 ⁻¹	301.0	329.0	31.8
225	1.4X10 ⁻⁴	58.4	77.0	215.0
	5.6X10 ⁻⁴	82.0	104.0	141.0
	1.4X10 ⁻³	101.0	134.0	116.0
	5.6X10 ⁻³	124.0	167.0	140.0
	1.4X10 ⁻²	143.0	177.0	138.0
	5.6X10 ⁻²	216.0	253.0	99.2
	1.4X10 ⁻¹	262.0	280.0	37.3
250	1.4X10 ⁻⁴	28.4	36.0	269.0
	2.8X10 ⁻⁴	37.7	42.0	294.0
	5.6X10 ⁻⁴	46.1	59.0	335.0
	1.4X10 ⁻³	59.2	78.0	228.0
	2.8X10 ⁻³	71.0	91.0	135.0
	5.6X10 ⁻³	86.9	108.0	179.0
	1.4X10 ⁻²	105.0	134.0	142.0
	2.8X10 ⁻²	136.0	170.0	104.0
	5.6X10 ⁻²	170.0	206.0	121.0
	1.4X10 ⁻¹	191.0	218.0	54.8

*Specimen fractured before achieving 0.1 strain.

Temperature C	Strain Rate s ⁻¹	UTS Mpa	True Stress at 0.1 Plastic Strain		Ductility
			Mpa		
275	1.4X10 ⁻⁴	17.5	22.0	198.0	
	2.8X10 ⁻⁴	25.0	32.0	438.0	
	5.6X10 ⁻⁴	29.4	35.0	397.0	
	1.4X10 ⁻³	39.4	52.0	255.0	
	2.8X10 ⁻³	45.9	58.0	239.0	
	5.6X10 ⁻³	57.1	75.0	120.0	
	1.4X10 ⁻²	84.1	104.0	281.0	
	2.8X10 ⁻²	110.0	141.0	209.0	
	5.6X10 ⁻²	128.0	167.0	182.0	
1.4X10 ⁻¹	154.0	187.0	73.3		
300	1.4X10 ⁻⁴	11.4	14.0	258.0	
	2.8X10 ⁻⁴	12.9	15.0	283.0	
	5.6X10 ⁻⁴	20.1	23.0	392.0	
	1.4X10 ⁻³	25.9	32.0	391.0	
	2.8X10 ⁻³	28.9	39.0	373.0	
	5.6X10 ⁻³	48.1	60.0	293.0	
	1.4X10 ⁻²	61.3	83.0	160.0	
	2.8X10 ⁻²	77.9	94.0	238.0	
	5.6X10 ⁻²	93.8	111.0	138.0	
1.4X10 ⁻¹	120.0	154.0	85.2		
325	1.4X10 ⁻⁴	11.3	13.0	398.0	
	5.6X10 ⁻⁴	16.0	19.0	492.0	
	1.4X10 ⁻³	19.7	25.0	579.0	
	5.6X10 ⁻³	31.4	40.0	398.0	
	1.4X10 ⁻²	56.6	70.0	282.0	
	5.6X10 ⁻²	83.4	92.0	269.0	
	1.4X10 ⁻¹	108.0	137.0	187.0	
350	5.6X10 ⁻⁴	14.2	18.0	362.0	
	5.6X10 ⁻³	36.3	43.0	319.0	
	5.6X10 ⁻²	82.7	83.0	168.0	
375	5.6X10 ⁻⁴	13.7	16.0	498.0	
	5.6X10 ⁻³	40.1	45.0	191.0	
	5.6X10 ⁻²	78.6	77.0	216.0	
400	5.6X10 ⁻⁴	11.1	13.0	539.0	
	5.6X10 ⁻³	26.5	32.0	441.0	
	5.6X10 ⁻²	64.6	64.0	157.0	
425	5.6X10 ⁻⁴	5.9	7.0	684.0	
	5.6X10 ⁻³	16.5	22.0	326.0	
	5.6X10 ⁻²	41.7	44.0	150.0	

STRESS VS STRAIN

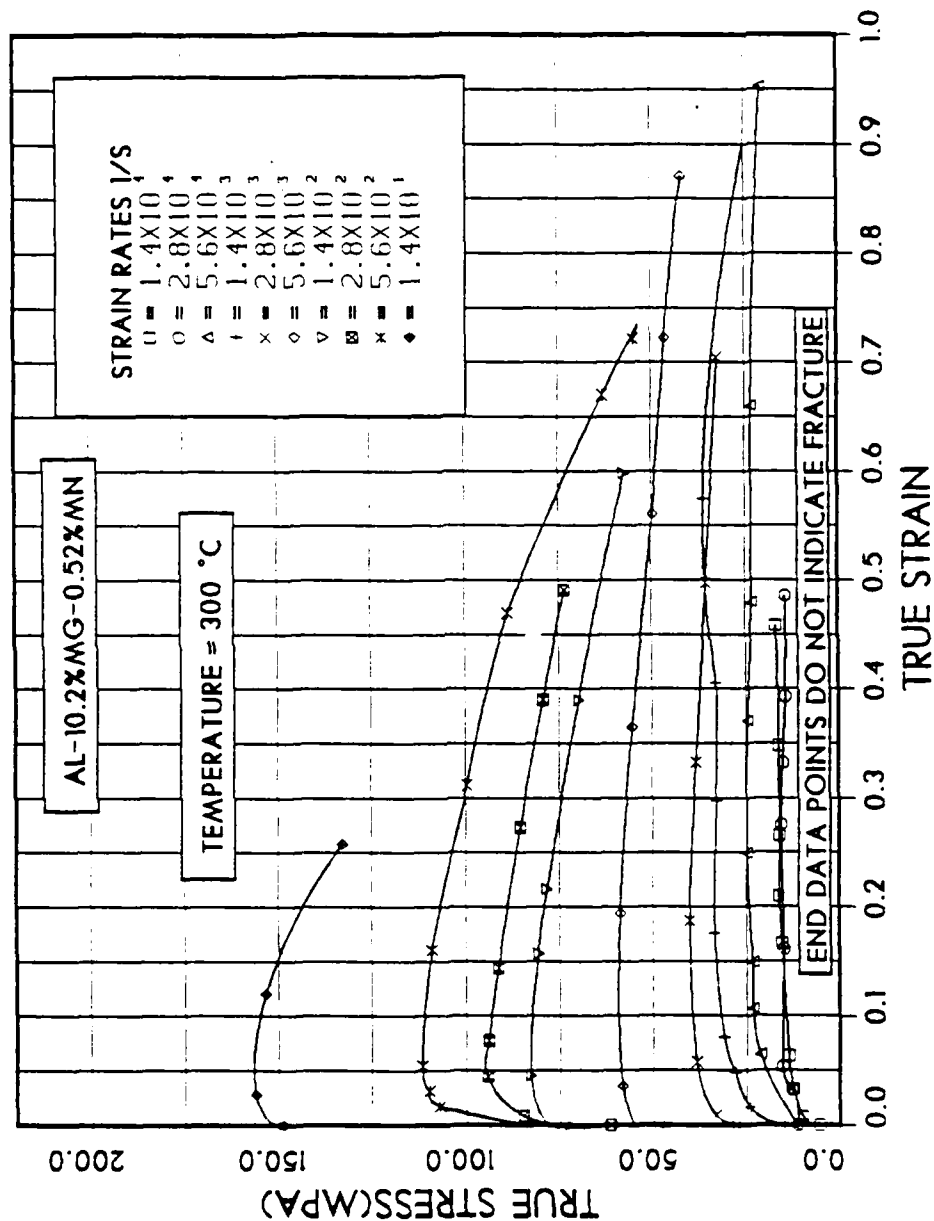


Figure 4.1 True Stress vs True Plastic Strain Data for Testing Conducted at 300 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

increasing strain. Such an apparent softening could result from localized deformation of the samples. The softening, however, generally appears at high strain rates. Jonas [Ref. 13] reported similar data and suggested that this was due to a break up of a fibered structure resulting from rolling. A detailed explanation as to why this should result was not offered although it may be inferred that the more equiaxed structure had a smaller apparent grain size. At intermediate strain rates the stress remains relatively constant over a wide strain range, and at low strain rates the stress increases slightly from strain hardening, perhaps due to grain growth.

This latter behavior can be understood from models such as those due to Nabarro [Ref. 14] and Herring [Ref. 15], or Coble [Ref. 16], all of which predict $1/d^x$ grain size dependence for the deformation rate, where d = grain size and $x = 2$ (Nabarro-Herring) or 3 (Coble). As grain growth occurs, the stress must increase to maintain a constant strain rate. To obtain representations for the temperature and strain rate dependence of plastic deformation, values of true stress at a true plastic strain of 0.1 were plotted against temperature (for each strain rate) and strain rate (for each temperature).

Figure 1.2 illustrates the dependence of the flow stress at 0.1 strain on temperature for each strain rate examined. Generally, as the strain rate increases the

STRESS VS TEMPERATURE

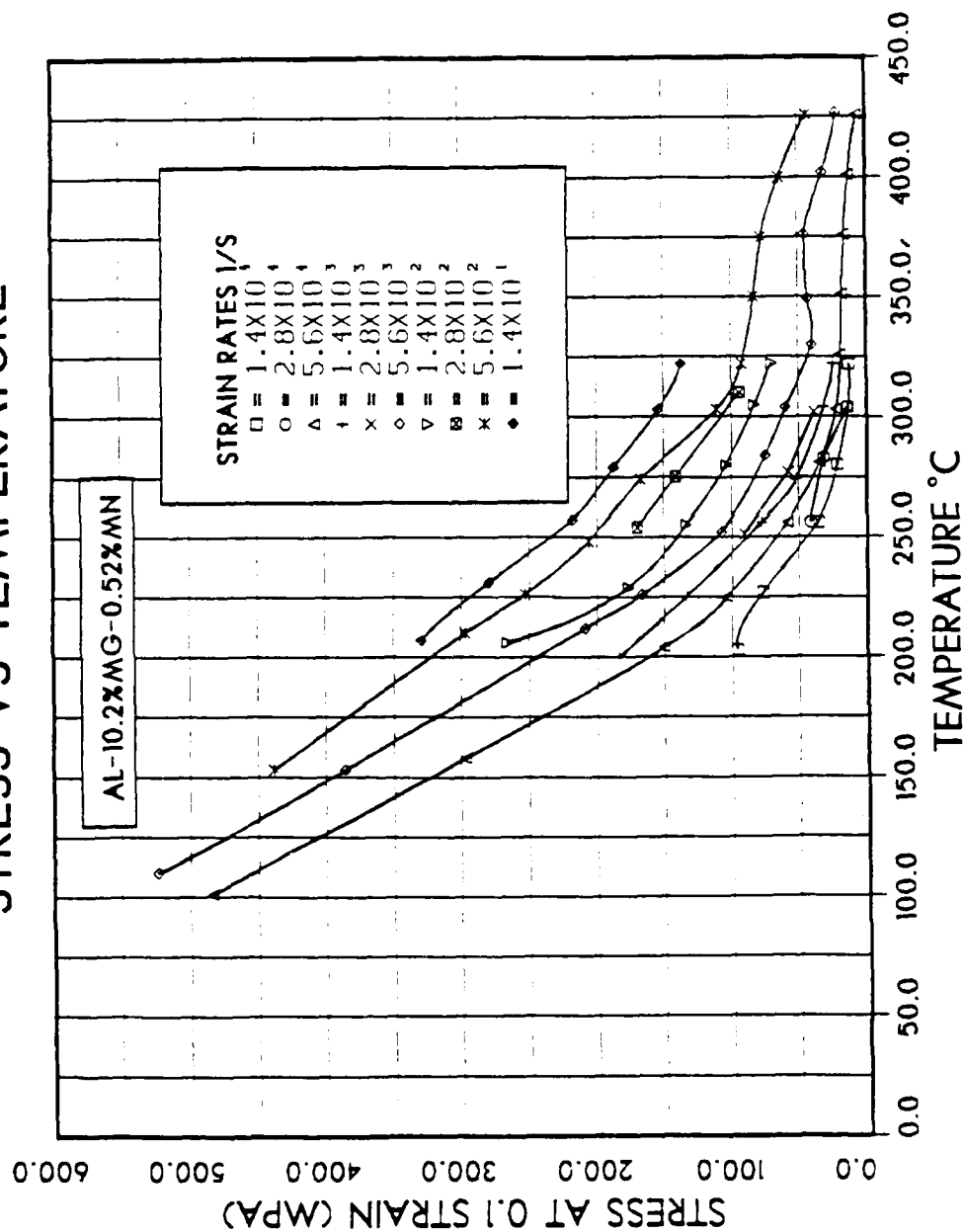


Figure 4.2 True Stress vs Temperature data for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 hours, Annealed at 440 C for 1 hour, Oil quenched, and Warm Rolled at 300 C to 94% Reduction.

stress increases and as temperature increases the stress decreases. The trend of the curves suggests a weakening of the temperature dependence of the flow stress for temperatures above the rolling temperature, 300 C.

Sherby et. al. [Ref. 17] have noted that one common characteristic of superplastic metallic alloys was that their resistance to plastic flow is highly strain rate sensitive. Figures 4.3 and 4.4 are plots of log stress at 0.1 plastic strain vs log strain rate for selected temperatures. The data of Figures 4.3 and 4.4 are generally not linear for each temperature. Rather, the slope m generally increases with decreasing strain rate, although at temperatures from 275 C to 325 C the curves appear sigmoidal as discussed by Mukherjee [Ref. 18]. Also, the slopes appear to increase with increasing temperature for any strain rate. The data in Figure 4.4 was plotted separately to avoid overlap; as noted previously, the flow stress dependence upon temperature is weak in this range.

Based on the stress-temperature data of Figure 4.2, activation energies can be determined by plotting strain rate vs. $1/T$ at constant values of stress (Figure 4.5). These activation energy values were obtained from the data of Figure 4.5 by applying the relation

$$Q = -R \frac{\ln \dot{\epsilon}}{1/T} \quad (\dot{\epsilon} = \text{CONSTANT}) \quad (\text{eqn. 4.1})$$

STRESS VS STRAIN RATE

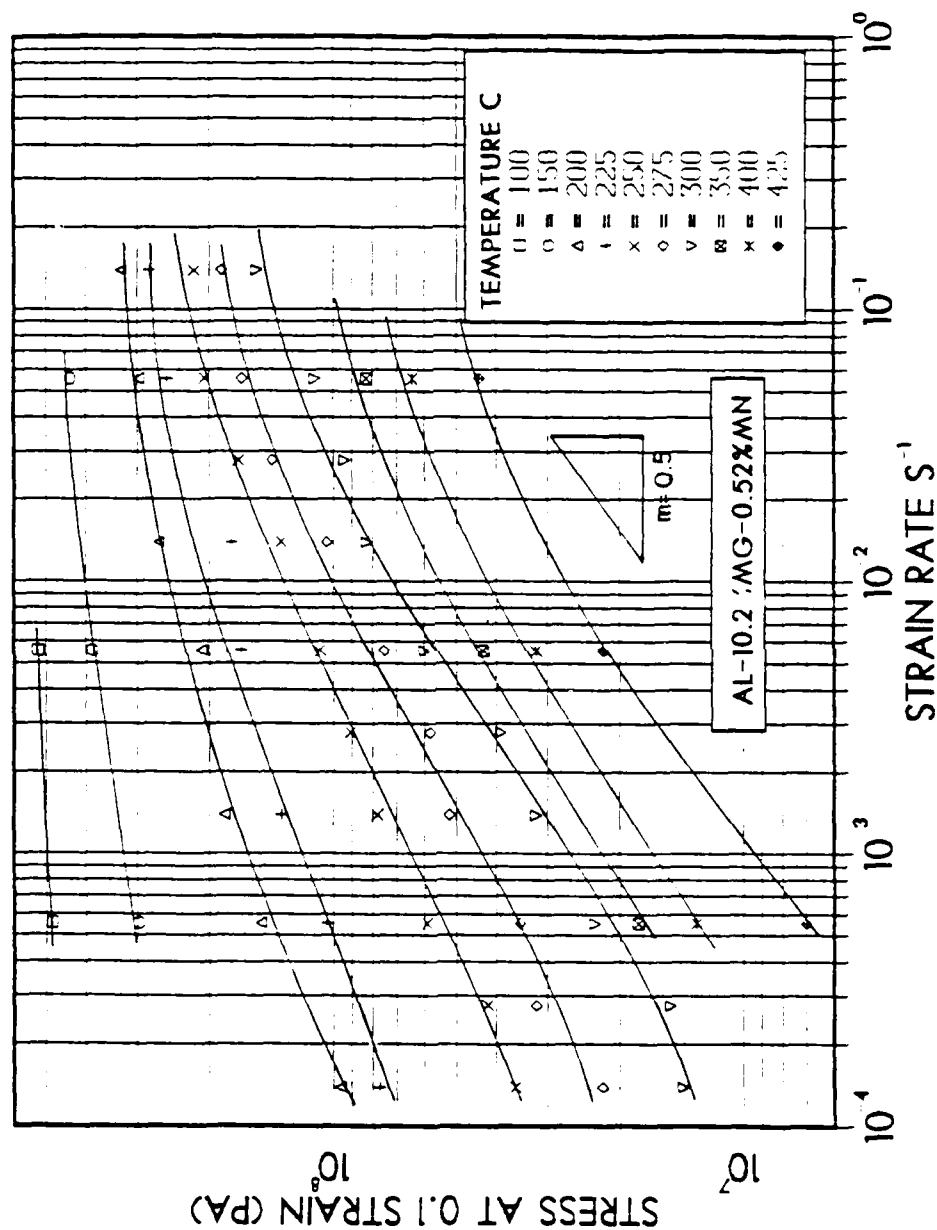


Figure 4.3 True Stress vs Strain Rate Data for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 hours, Annealed at 440 C for 1 hour, Oil Quenched, and Warm Rolled at 300 C to 94% Reduction.

STRESS VS STRAIN RATE

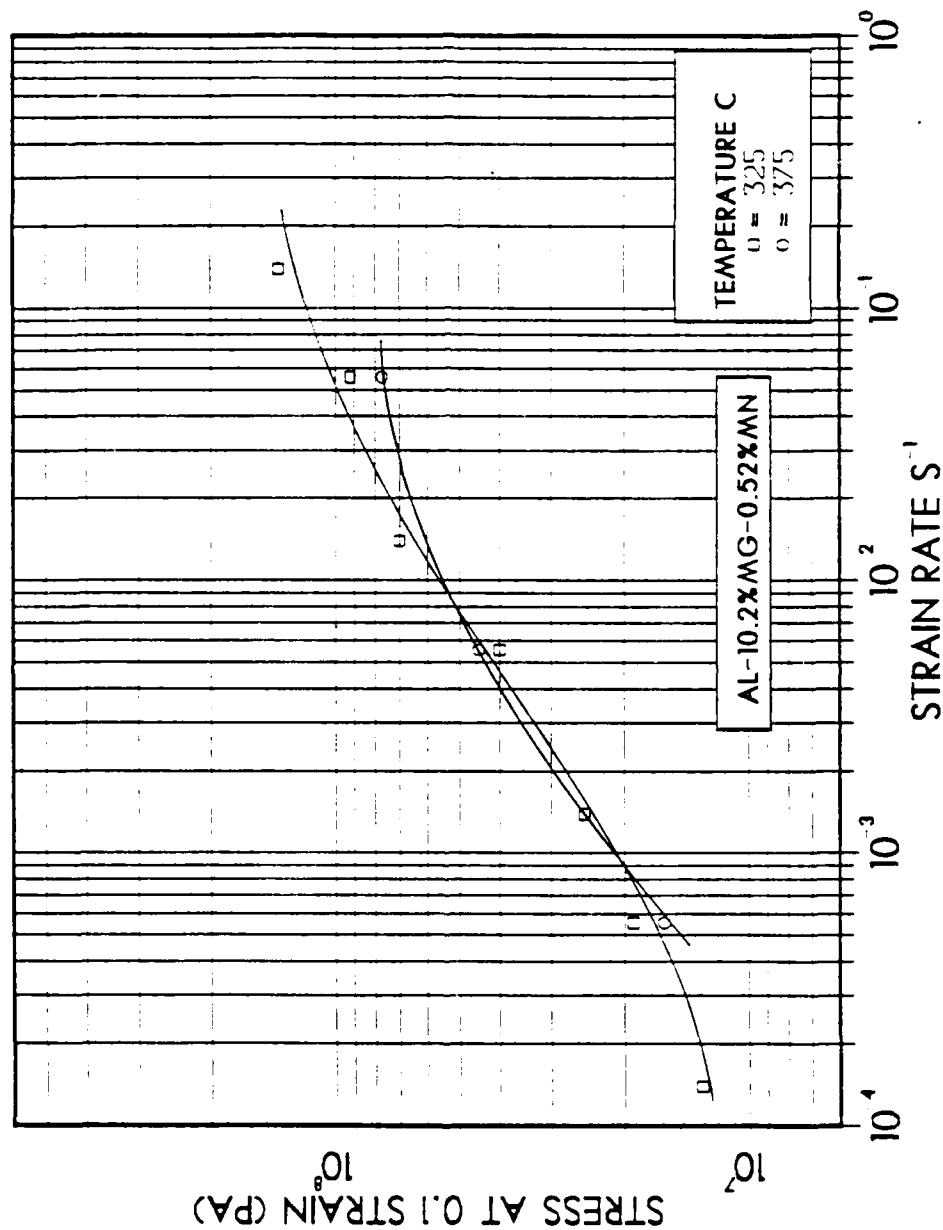


Figure 4.4 True Stress at 0.1 Strain vs Strain Rate Data for Testing, Conducted at Temperatures of 325 C and 375 C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 Hours, Annealed at 440 C for 1 Hour, Oil Quenched, and Warm Rolled at 300 C to 94% Reduction.

STRAIN RATE VS 1/T

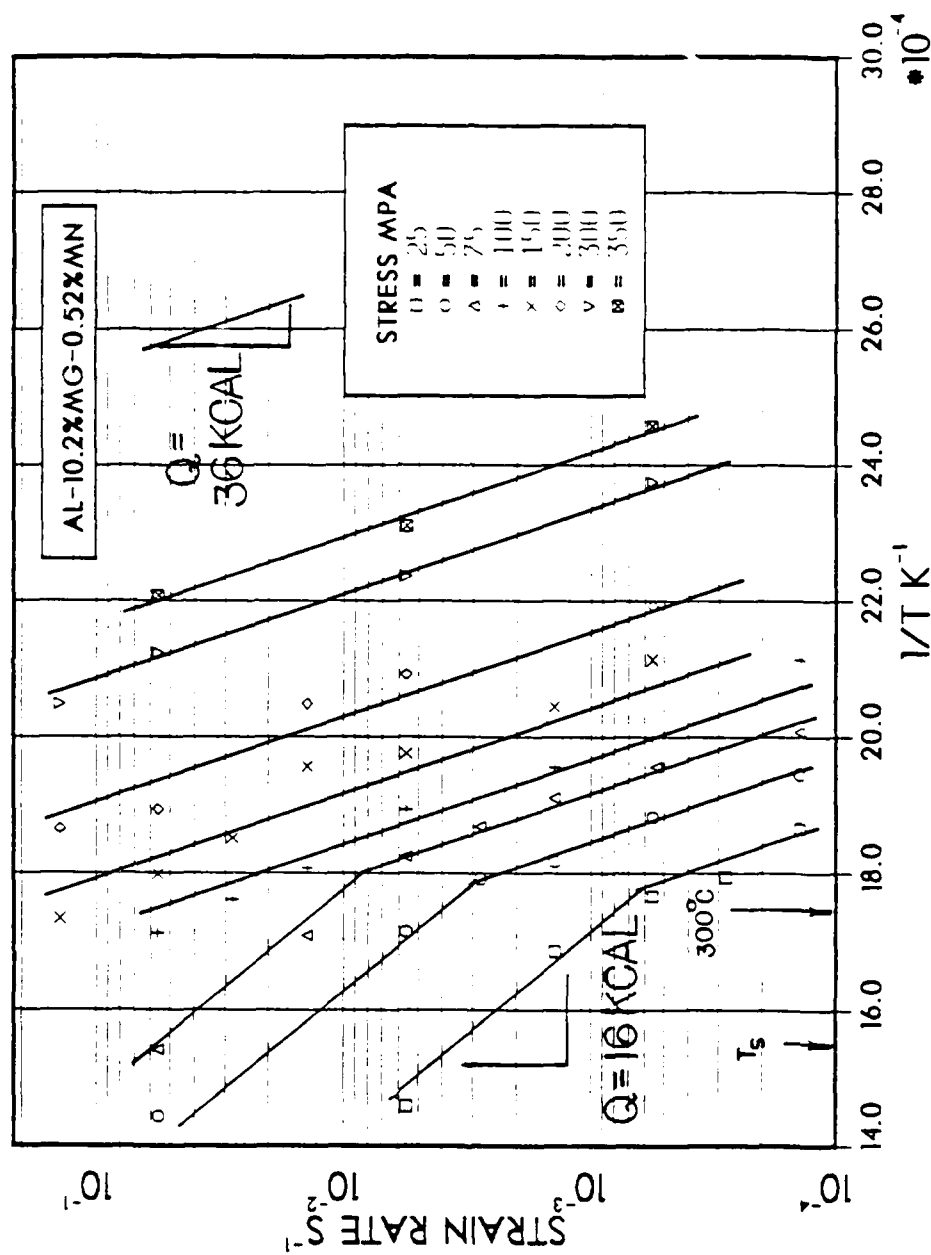


Figure 4.5 Strain Rate vs 1/T Data for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 hours, Annealed at 440 C for 1 hour, Oil Quenched, and Warm Rolled at 300 C to 94% Reduction.

for each of the stresses indicated; that is, the slopes of the individual lines on Figure 4.5 were used to obtain the values shown.

The activation energy at higher stresses and lower temperatures is about 36 Kcal/mol. This value is consistent with lattice diffusion control of deformation, either via control of dislocation climb or dislocation glide [Ref. 20]. A change in slope appears to occur near 300 C; above this temperature (at smaller values of $1/T$) the activation energy appears to decrease to a value of about 16 Kcal/mol. The rate and the temperatures at which this value becomes dominant correspond to the regime wherein the apparent temperature dependence of the stress diminishes (Figure 4.2). The rates and temperatures also correspond to those wherein superplastic ductilities begin to be observed.

Micrographs in Figures 4.6 and 4.7 are of samples tested at 200 C and 300 C, respectively. No cavitation appears at 200 C although some cavitation is observed at 300 C. These observations are consistent with the noted break in slope in Figure 4.5 at about 300 C indicating the onset of possible grain boundary sliding.

At temperatures above the solvus (367 C) the magnesium tends to go back into solution, with the result that the intermetallic is no longer present to retard coarsening of the subgrain structure or to inhibit recrystallization. The recrystallization coupled with the solid solution

Al-10Mg-0.52Mn

$\dot{\epsilon} = 5 \times 10^{-4} \text{ SEC}$ 200°C

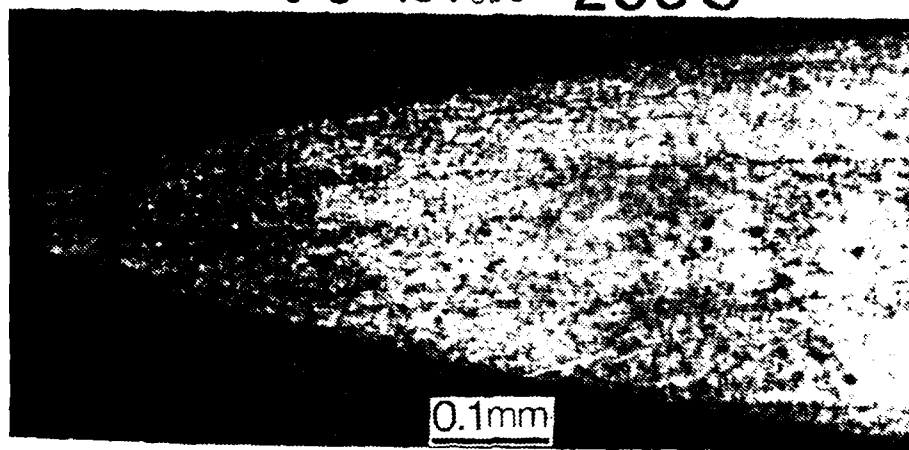


Figure 4.6 Optical Micrograph of Al-10.2%Mg-0.52%Mn, 160x, Tested at 200 C, Strain Rate $5.6 \times 10^{-4} \text{ s}^{-1}$; Sectioned Longitudinally. Etched Using Graf-Sargent Solution.

Al-10Mg-0.52Mn

$\dot{\epsilon} = 5 \times 10^{-4} \text{ s}^{-1}$ 300°C

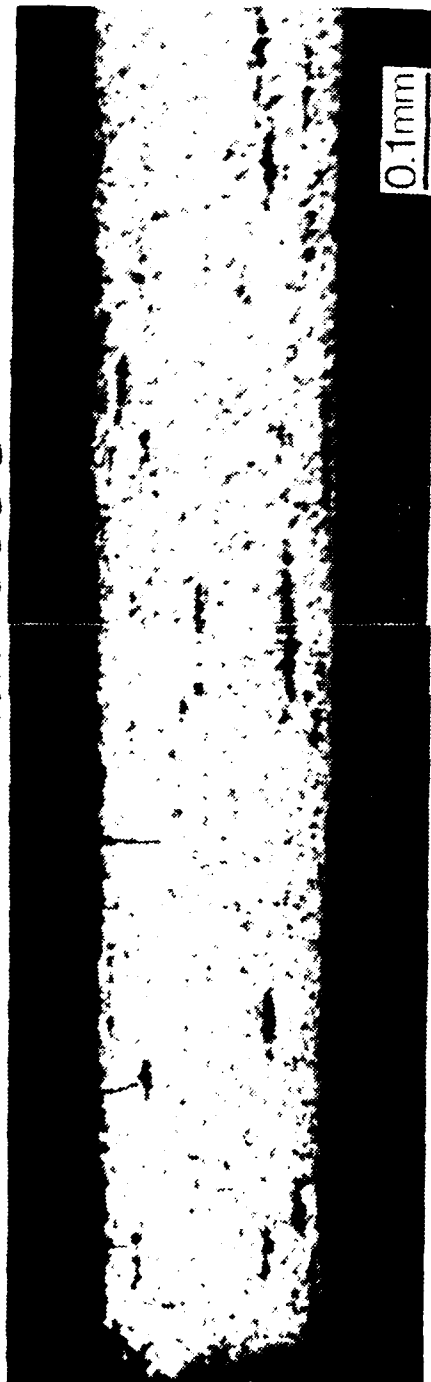


Figure 4.7 Optical Micrograph of Al-10.2%Mg-0.52%Mn, 160X, Tested at 300 °C, Strain Rate $5.6 \times 10^{-4} \text{ s}^{-1}$; Sectioned Longitudinally to Reveal Cavitation. Etched Using Graf-Sargent Solution.

strengthening within the lattice may promote grain boundary sliding as the dominant deformation mechanism. A characteristic of recrystallized aluminum alloys undergoing superplastic deformation via grain boundary sliding is extensive cavitation. The micrograph in Figure 4.8 shows extensive cavitation in the test specimen pulled at 400 C.

Ductility was plotted versus temperature in Figure 4.9 for the warm rolled Al-10.2%Mg-0.52%Mn alloy of this research. Included is data from Becker [Ref. 10] and Stengel [Ref. 19] on this alloy, warm rolled and then recrystallized by annealing at 440 C prior to tension testing. The data on the material recrystallized represents a pattern expected for these aluminum alloys. The as-rolled data, however, rises significantly in ductility between 150 C and 300 C. Sample elongations of greater than 400% were observed at temperatures as low as 275 C. This indicates that warm rolling enhances ductility to values greater than expected. Theories of elevated temperature deformation do not consider subgrain structures as likely to exhibit superplastic behavior. Rather, fine grain size is thought to be required. It is not clear, here, why such structures exhibit such enhanced ductility, but the ductility itself is clearly the result of the warm rolling. The increasing n value with increasing temperature also would result from the warm rolling. The

Al-10Mg-0.52Mn

400°C $\dot{\epsilon} = 5 \times 10^{-4} \text{ s}^{-1}$

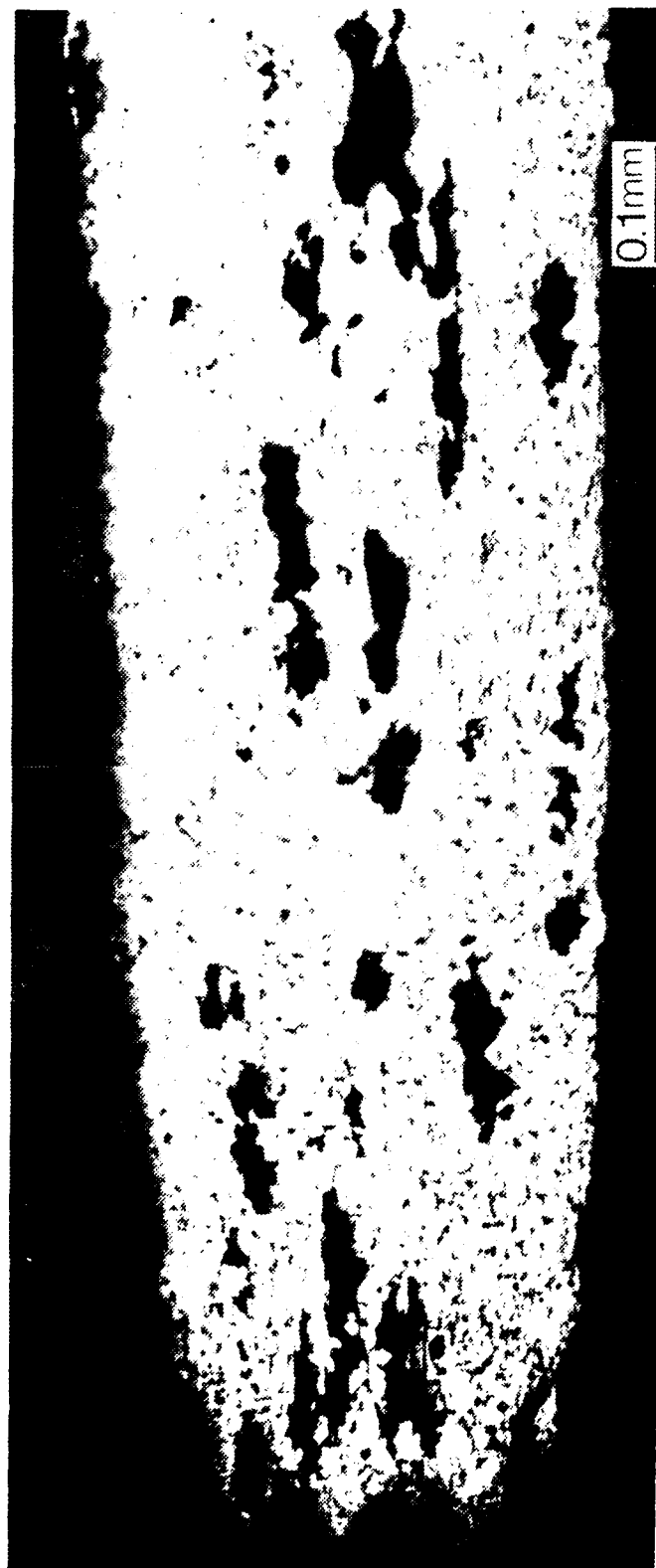


Figure 4.8 Optical Micrograph of Al-10.2%Mg-0.52%Mn, 160X, Tested at 400 C, Strain Rate $5.6 \times 10^{-4} \text{ s}^{-1}$; Sectioned Longitudinally to Reveal Cavitation. Etched Using Graf-Sargent Solution.

DUCTILITY VS TEMPERATURE

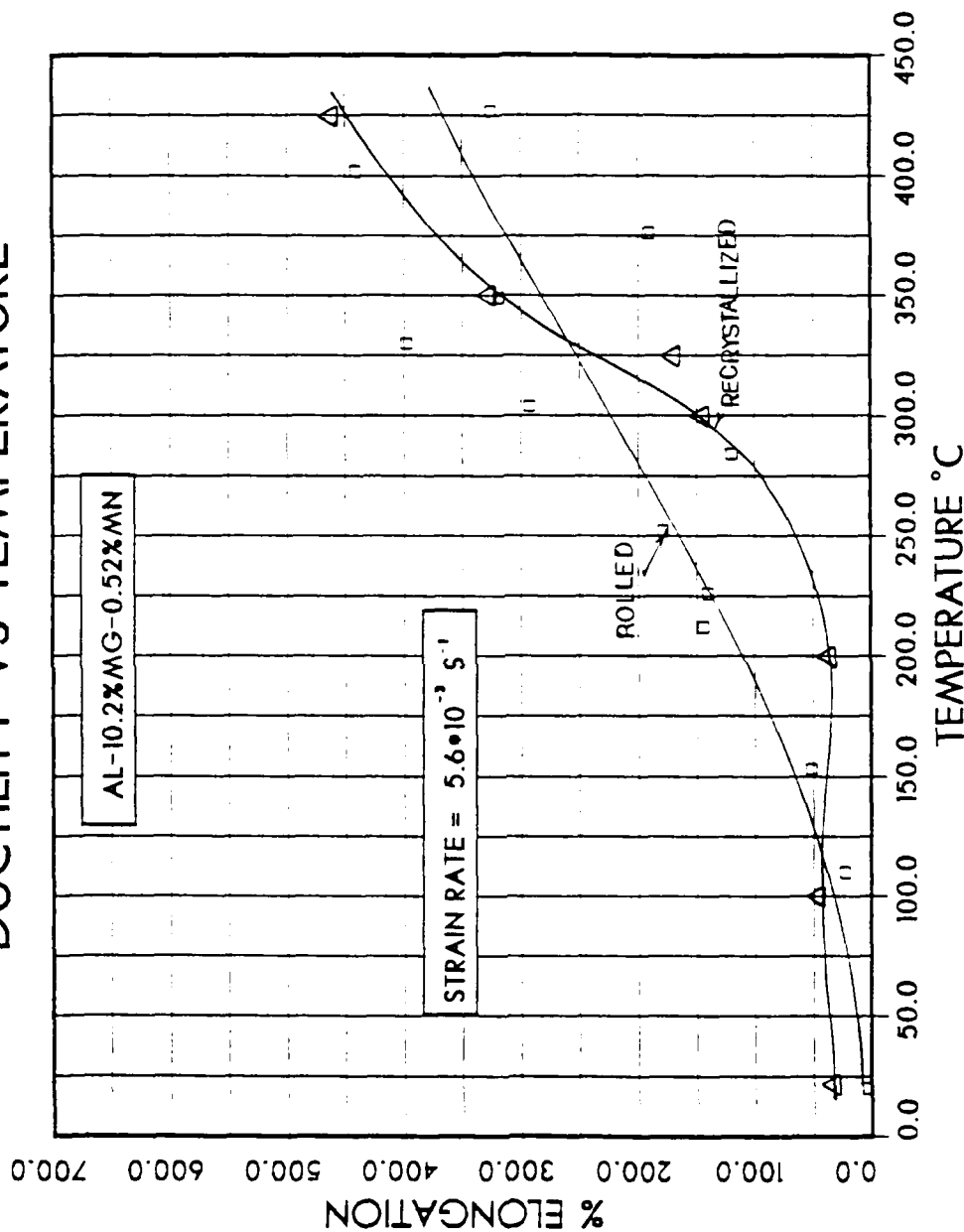


Figure 4.9 Ductility vs Temperature Data Comparing As-Rolled to Recrystallized Data for Testing Conducted at $5.6 \times 10^{-3} \text{ s}^{-1}$ for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours; Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

remaining plots and data are included in Appendix A and Table II.

Figure 4.10 is a plot of ductility vs strain rate for a constant temperature of 300 C. The curve describes an expected shape, based on the stress-strain rate data. It should be noted that peak ductility of 392% occurs at a strain rate of $5.6 \times 10^{-3} \text{ s}^{-1}$, a relatively high strain rate. More of these plots at selected temperatures are included in Appendix A.

B. METALLOGRAPHY

A comparison between microstructures after testing at strain rates of 5.6×10^{-2} and $5.6 \times 10^{-4} \text{ s}^{-1}$ can be seen in Figure 4.11. There is a marked difference in grain structures as a function of strain rate. At high rate where ductility is lower and time at temperature is short, little cavitation is seen and little evidence for resolution of the second phase or recrystallization. On the other hand, at a lower rate with more time at temperature, the resolution of the second phase and recrystallization lead to more ready boundary sliding and the accompanying cavitation.

In summary, the activation energy for deformation follows a pattern suggesting lattice diffusion giving way to grain boundary diffusion control as temperature increases above 300 C, the rolling temperature. The m values attained

DUCTILITY VS STRAIN RATE

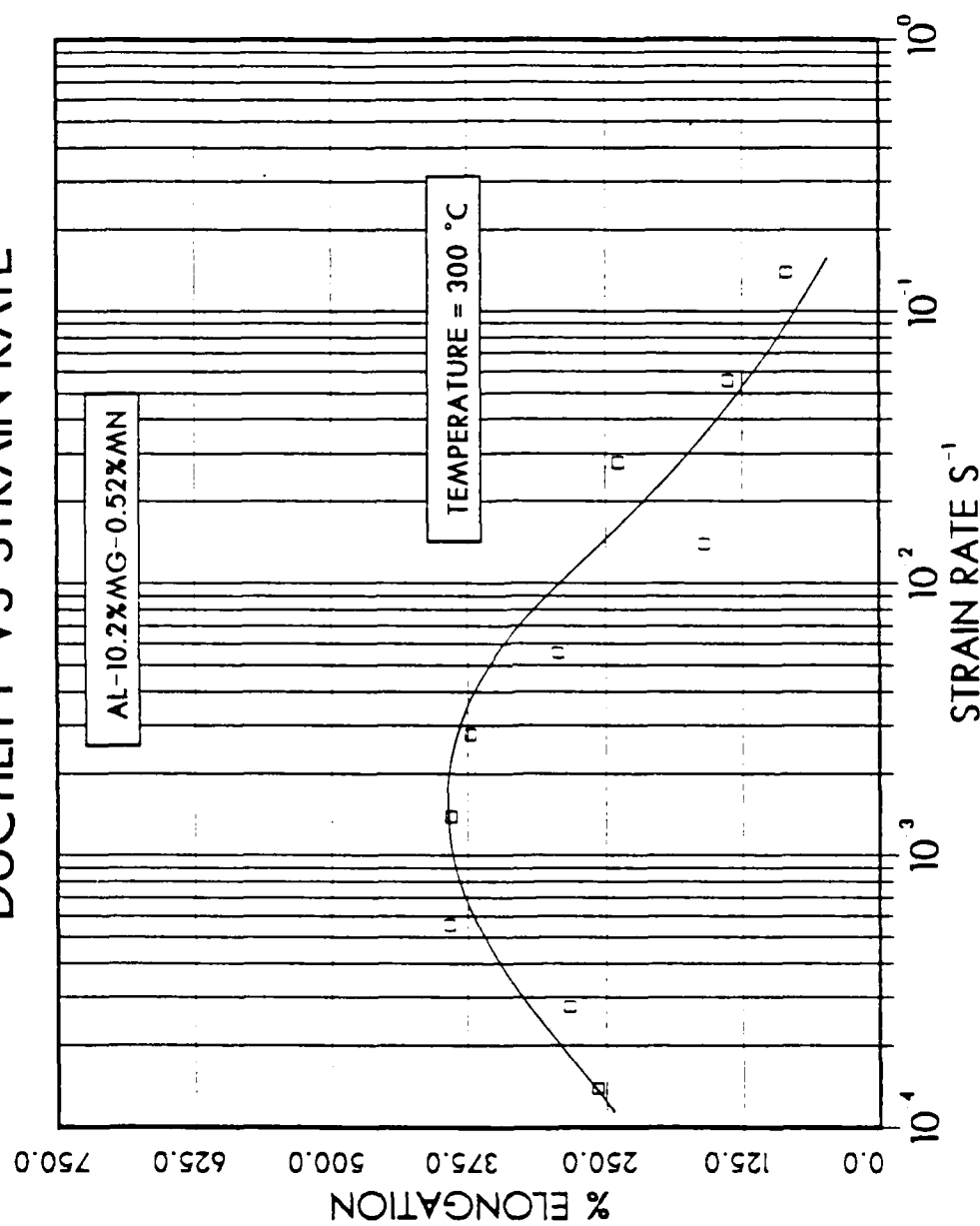
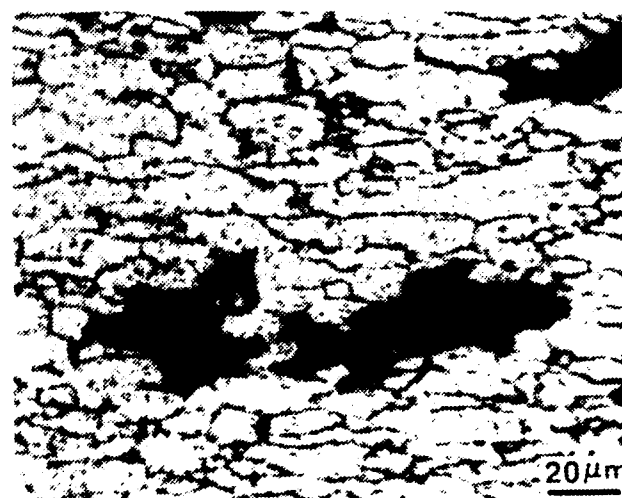


Figure 4.10 Ductility vs Strain Rate Data for Testing Conducted at 300 C for Al-10.2%Mg-0.52%Mn. Solution treated at 440 C for 24 hours, Annealed at 440 C for 1 hour, oil quenched, and Warm Rolled at 300 C to 94% Reduction.

Al-10Mg-0.52Mn
400°C



$\dot{\epsilon} = 5 \times 10^{-2} \text{ SEC}$



$\dot{\epsilon} = 5 \times 10^{-4} \text{ SEC}$

Figure 4.11 Optical Micrographs of Al-10.2%Mg-0.52%Mn, 500X, Tested at 400 C, Sectioned Longitudinally, to Compare Grain Size and Extent of Cavitation. Strain Rates Were $5 \times 10^{-2} \text{ s}^{-1}$ and $5 \times 10^{-4} \text{ s}^{-1}$, Respectively. Etched Using Graf-Sargent Solution.

were 0.4-0.5 at approximately 300 C and resulted in superplastic ductility in a structure consisting initially of fine subgrains rather than grains. These observations suggest further development of dislocation models is needed. Also, it appears that current grain boundary sliding models seem inadequate to explain the observed behavior.

V. CONCLUSIONS AND RECOMMENDATIONS

The conclusions drawn from this research are: 1) warm rolled Al-10.2%Mg-0.52%Mn alloy is superplastic at temperatures as low as 275C; 2) the warm rolled alloy exhibits elongations of 400% at 300 C and strain rates of $5 \times 10^{-3} \text{ s}^{-1}$; 3) the warm rolling is responsible for the superplastic response at lower temperatures (near 300 C); 4) grain boundary sliding appears to be the predominant superplastic deformation mechanism at higher temperatures (above 300 C), based upon activation energy data; 5) microstructural data indicates that the structure prior to testing consists principally of fine subgrains rather than grains.

Recommendations for further work are: 1) microstructural analysis be conducted to reconcile the observations of activation energies appropriate for boundary sliding with the observations of dislocation substructures being present; 2) investigation into alloying effects on microstructure and superplasticity; 3) examination and further analysis of microstructural effects of annealing and recrystallization in this alloy.

LIST OF REFERENCES

1. Ness, F. G., Jr., High Strength to Weight Aluminum-18 Weight Percent Magnesium Alloy Through Thermomechanical Processing, M.S. Thesis, Naval Postgraduate School, Monterey, California, December 1976.
2. Bingay, C. P., Microstructural Response of Aluminum-Magnesium Alloys to Thermomechanical Processing, M.S. Thesis, Naval Postgraduate School, Monterey, California, December 1977.
3. Glover, T. L., Effects of Thermomechanical Processing on Aluminum-Magnesium Alloys Containing High Weight Percent Magnesium, M.S. Thesis, Naval Postgraduate School, Monterey, California, December 1977.
4. Grandon, R. A., High Strength Aluminum-Magnesium Alloys: Thermomechanical Processing, Microstructure and Tensile Mechanical Properties, M.S. Thesis, Naval Postgraduate School, Monterey, California, December 1976.
5. Speed, W. G., An Investigation into the Influence of Thermomechanical Processing on Microstructure and Mechanical Properties of High Strength Aluminum-Magnesium Alloys, M.S. Thesis, Naval Postgraduate School, Monterey, California, December 1977.
6. Chesterman, C. W., Jr., Precipitation, Recovery and Recrystallization Under Static and Dynamic Conditions for High Magnesium Aluminum-Magnesium Alloys, M.S. Thesis, Naval Postgraduate School, Monterey, California, March 1980.
7. Johnson, R. B., The Influence of Alloy Composition and Thermomechanical Processing Procedure on Microstructural and Mechanical Properties of High-Magnesium Aluminum Magnesium Alloys, M.S. Thesis, Naval Postgraduate School, Monterey, California, June 1980.
8. Shirah, R. H., The Influence of Solution Time and Quench Rate on the Microstructure and Mechanical Properties of High Magnesium Aluminum-Magnesium Alloys, M.S. Thesis, Naval Postgraduate School, Monterey, California, December 1981.

9. McNelley, T. and Garg, A., "Development of Structure and Mechanical Properties in Al-10.2%Mg by Thermo-Mechanical Processing", unpublished research, Naval Postgraduate School, Monterey, California.
10. Becker, J. J., Superplasticity in Thermomechanically Processed High Magnesium Aluminum Magnesium Alloys, M.S. Thesis, Naval Postgraduate School, Monterey, California, March 1984.
11. Bly, D. C., Sherby, O. D. and Young, C. M., "Influence of Thermal Mechanical Treatments on the Mechanical Properties of a Finely Spheroidized Eutectic Composition Steel", Material Science and Engineering, V. 2, pp. 41-46, 1973.
12. Brick, Pense, and Gordon, Structure and Properties of Engineering Materials, McGraw-Hill (Publishers) 1977.
13. Jonas, J. J., "Implications of Flow Hardening and Flow Softening During Superplastic Forming", Superplastic Forming of Structural Alloys, p. 57, 1982.
14. Nabarro, F. R. M., "Report of a Conference on the Strength of Solids", Physical Society, London, p. 75, 1948.
15. Herring, C., Journal of Applied Physics, 21, p. 437, 1950.
16. Coble, R. L., Journal of Applied Physics, 34, p. 1679, 1963.
17. Sherby, O. D. and Wadsworth, J., "Development and Characterization of Fine-Grain Superplastic Materials", Deformation, Processing, and Structure, pp. 354-384, 1982.
18. Mukherjee, A. K., "Deformation Mechanisms", Annual Review in Materials Science, V. 9, pp. 191-217, 1979.
19. Stengel, A., private communication, September 1984.
20. Sherby, O. D. and Burke, P. M., "Mechanical Behavior of Crystalline Solids at Elevated Temperature", Progress in Materials Science, V. 13, p. 325, 1968.

APPENDIX A

STRESS VS STRAIN

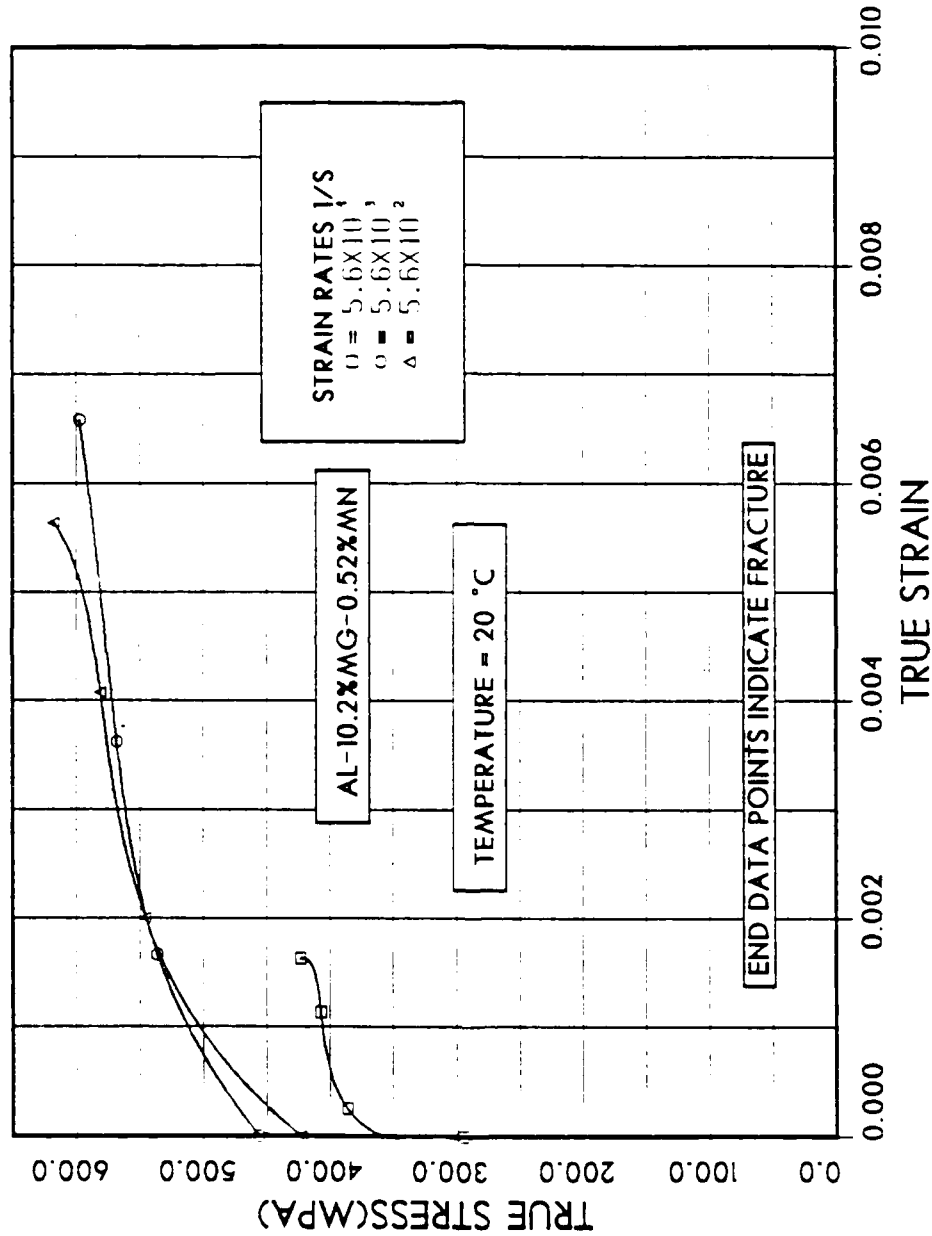


Figure A.1 True Stress vs True Plastic Strain Data for Testing Conducted at 20 °C for Al-10.2%Mg-0.5%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

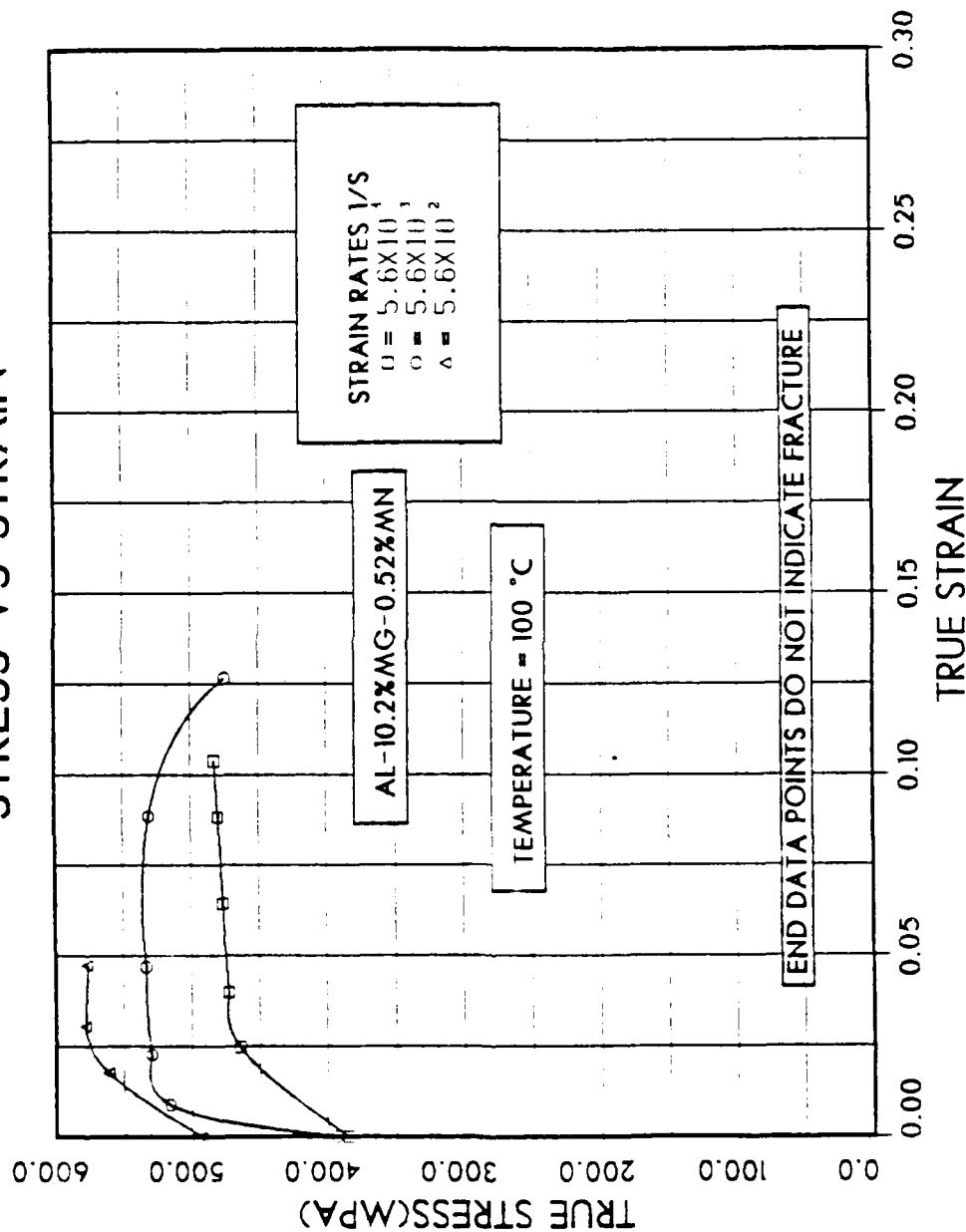


Figure A.2 True Stress vs True Plastic Strain Data for Testing Conducted at 100 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 hours, Annealed at 440 °C for 1 hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

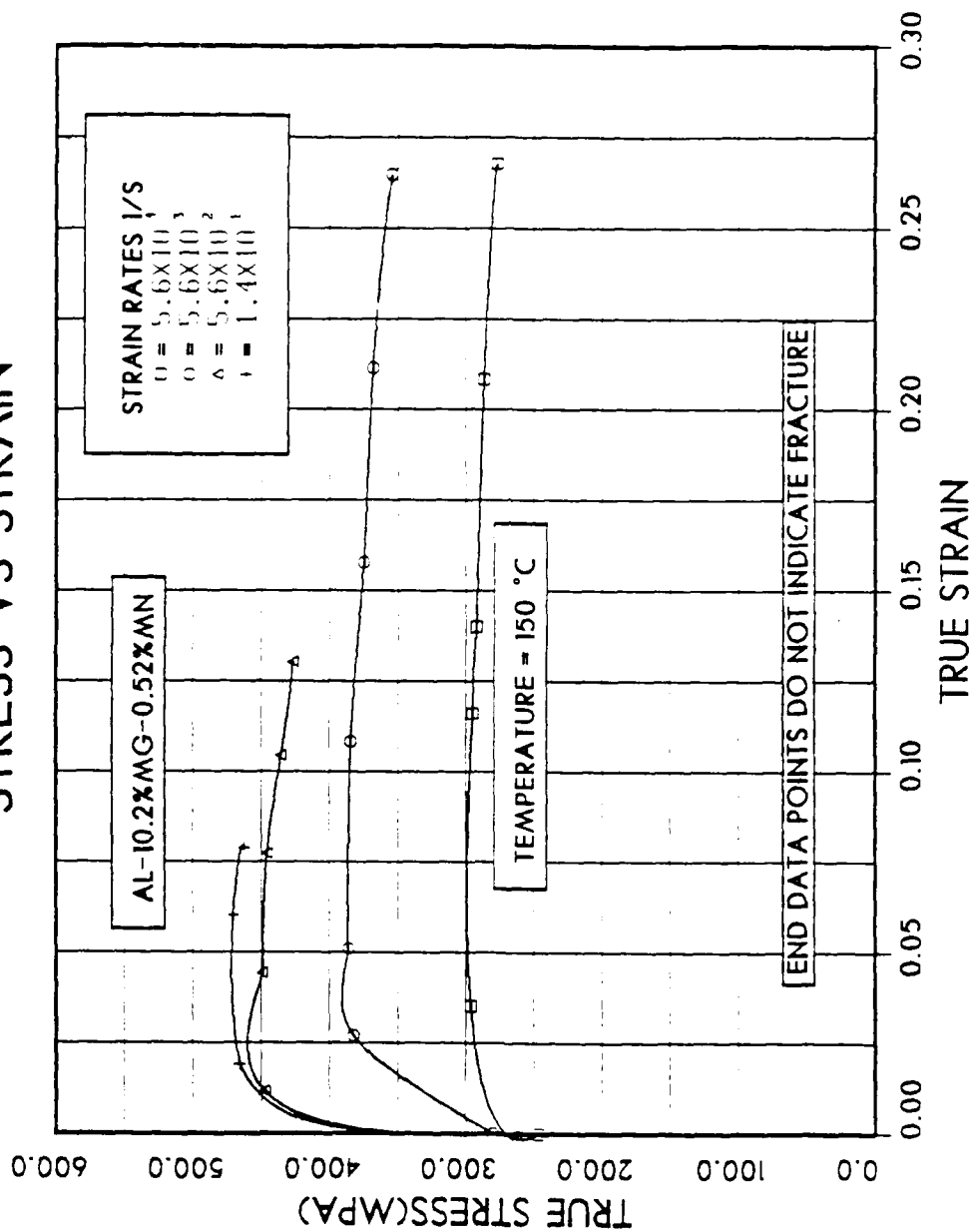


Figure A.3 True Stress vs True Plastic Strain Data for Testing Conducted at 150 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

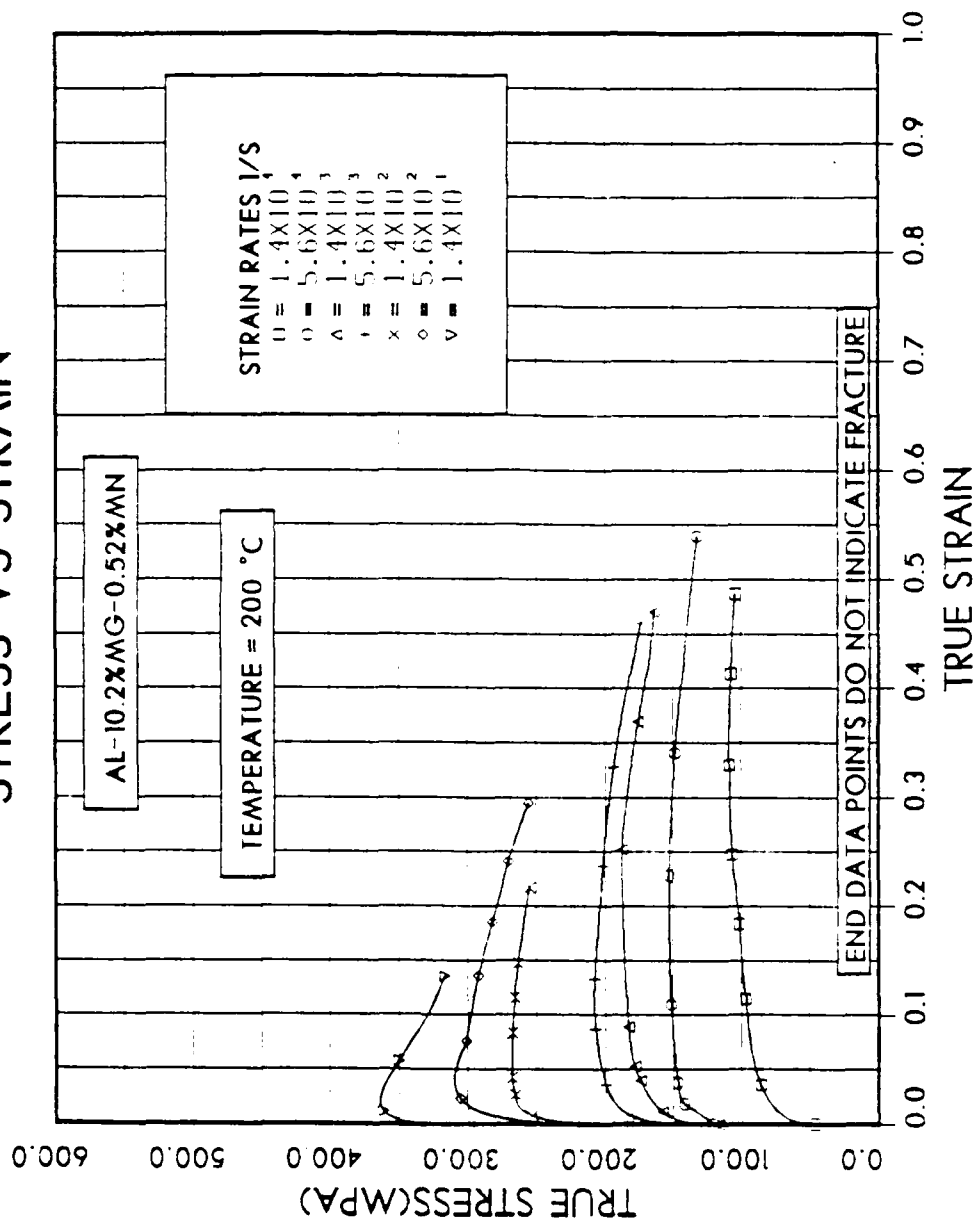


Figure A.4 True Stress vs True Plastic Strain Data for Testing Conducted at 200 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 hours, Annealed at 440 °C for 1 hour, oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

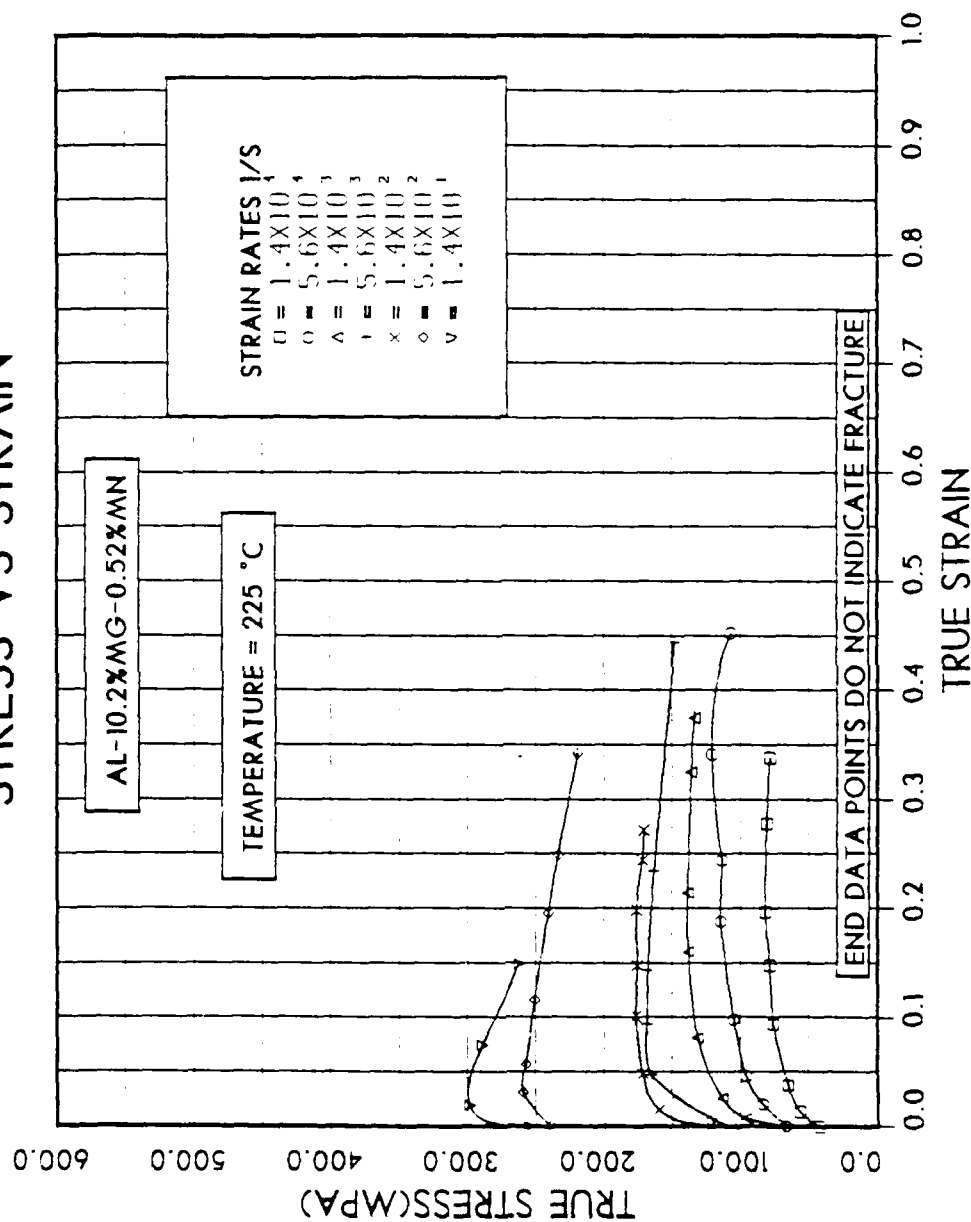


Figure A.5 True Stress vs True Plastic Strain Data for Testing Conducted at 225 °C for AL-10.2%MG-0.52%MN. Solution Treated at 440 °C for 24 hours, Annealed at 440 °C for 1 hour, oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

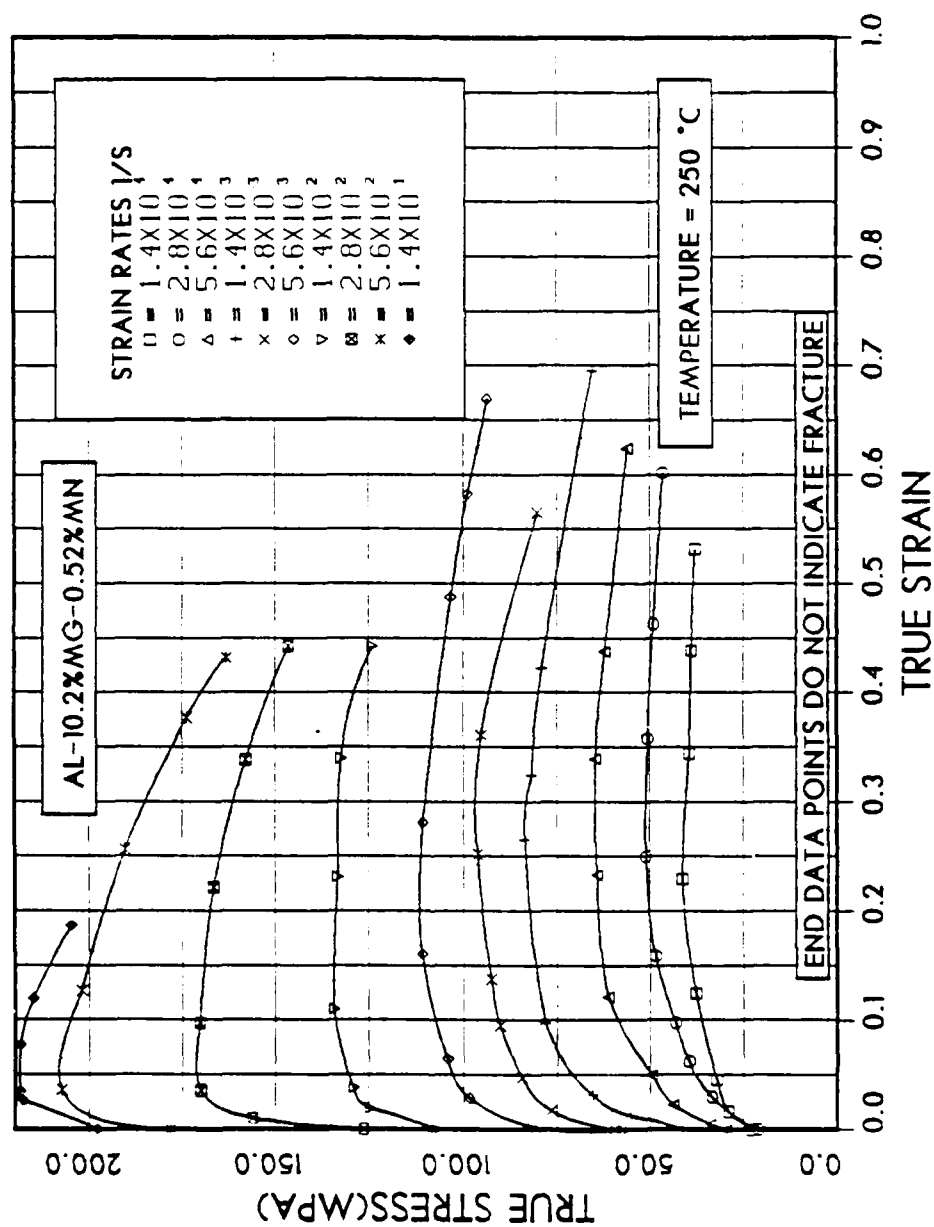


Figure A.6 True Stress vs True Plastic Strain Data for Testing Conducted at 250 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 hours, Annealed at 440 °C for 1 hour, oil quenched, and warm Rolled at 300 °C to 94% reduction.

STRESS VS STRAIN

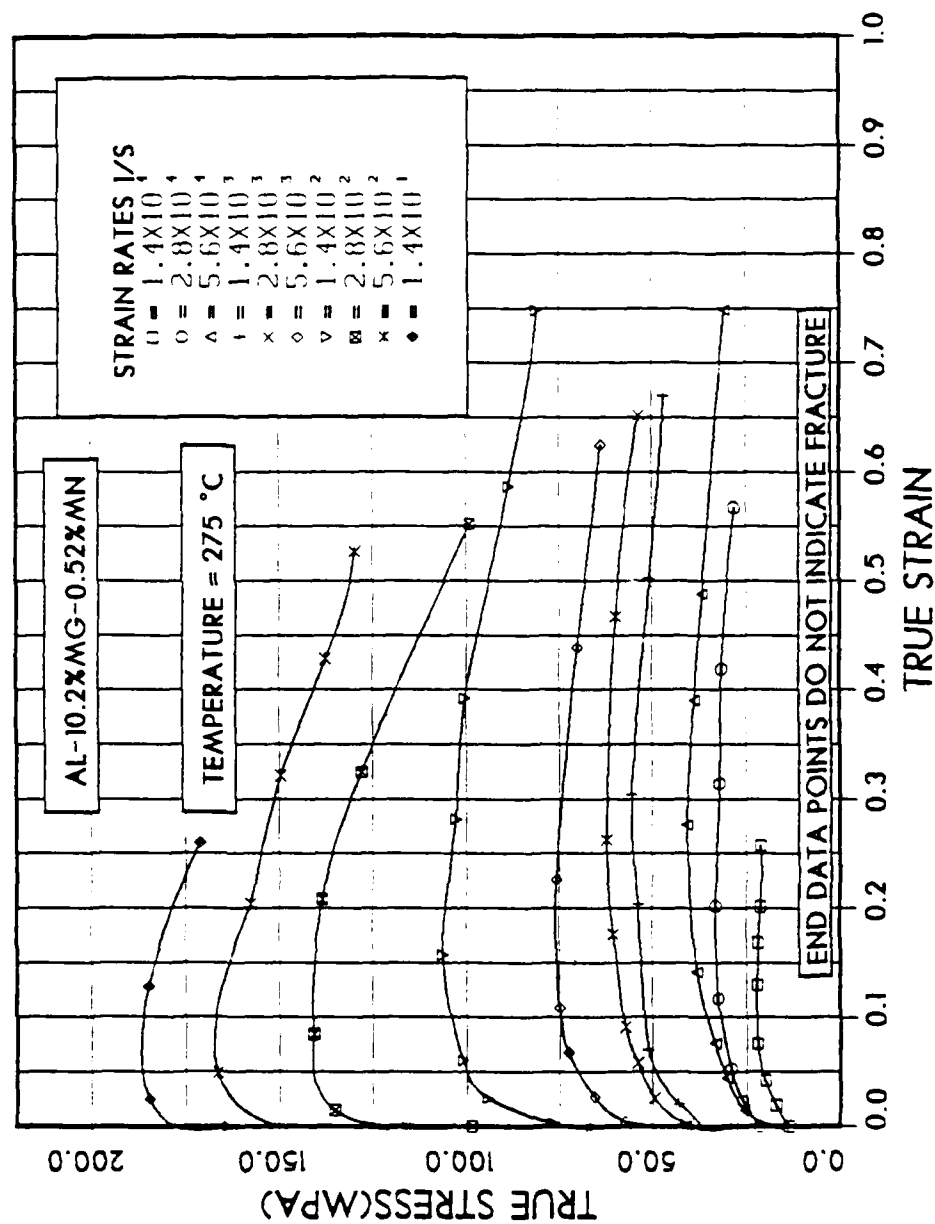


Figure A.7 True Stress vs True Plastic Strain Data for Testing Conducted at 275 C for AL-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 Hours, Annealed at 440 C for 1 hour, Oil Quenched, and Warm Rolled at 300 C to 94% Reduction.

STRESS VS STRAIN

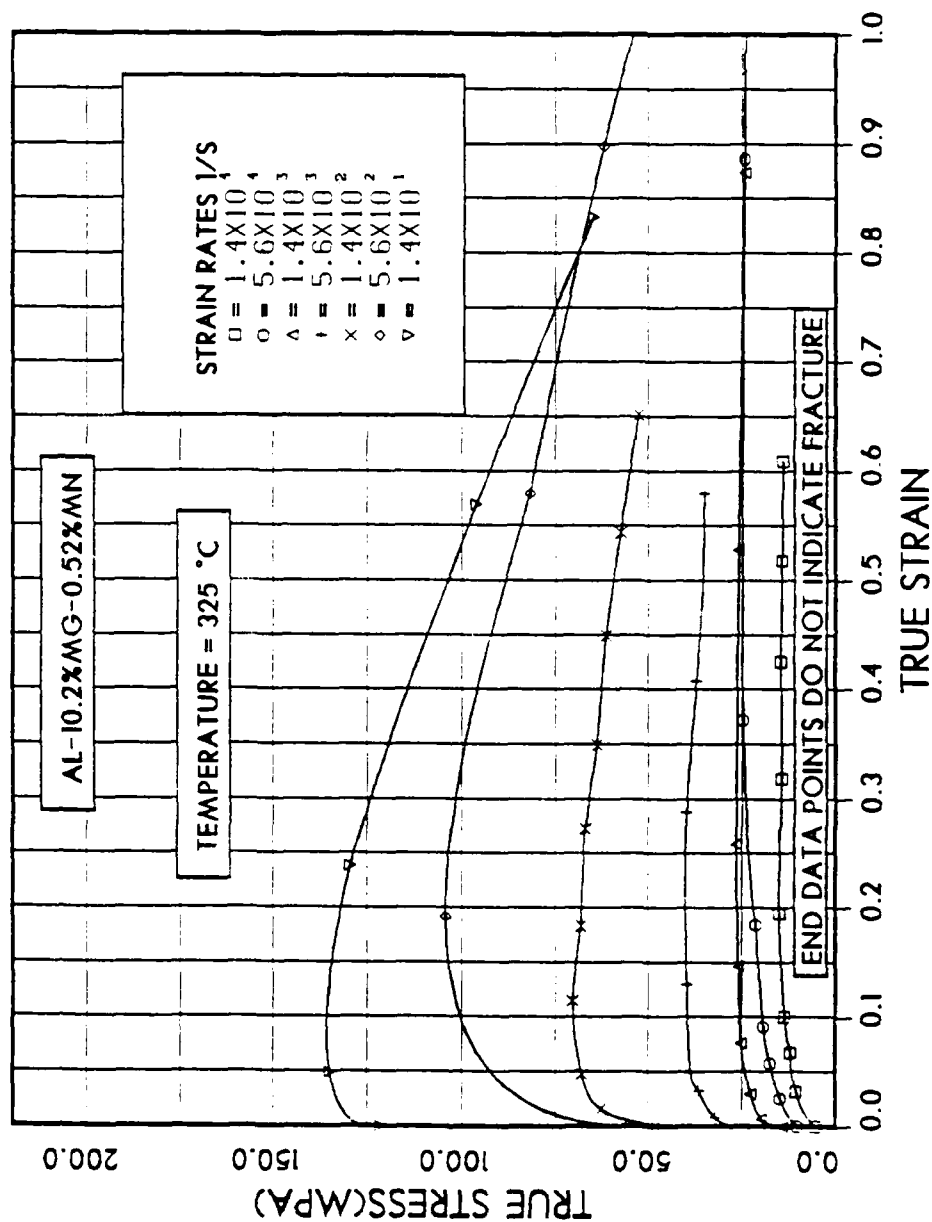


Figure A.8 True Stress vs True Plastic Strain Data for Testing Conducted at 325 °C for Al-10.2%Mg-0.52 Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

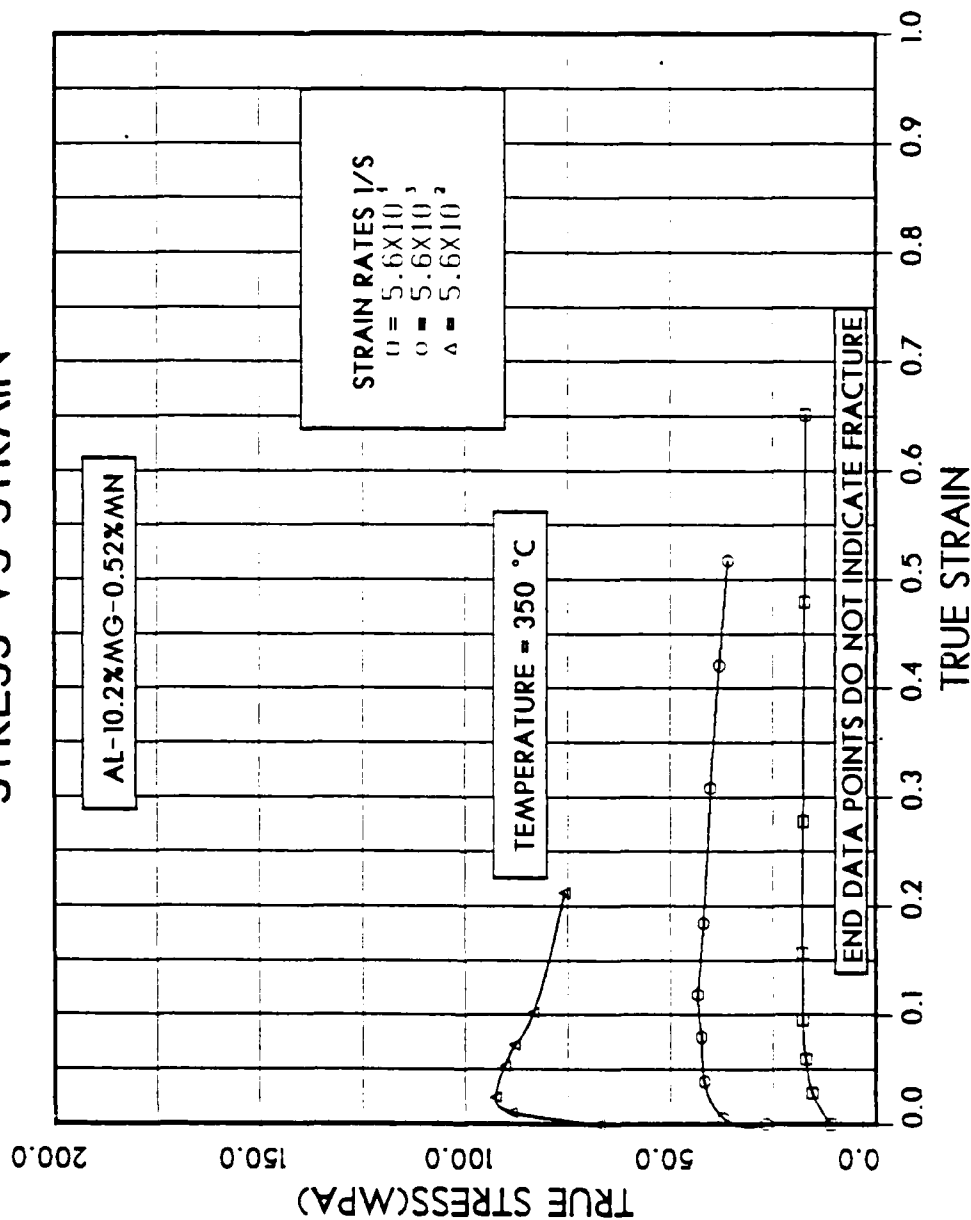


Figure A.9 True Stress vs True Plastic Strain Data for Testing Conducted at 350 C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 Hours, Annealed at 440 C for 1 Hour, Oil quenched, and Warm Rolled at 300 C to 94% Reduction.

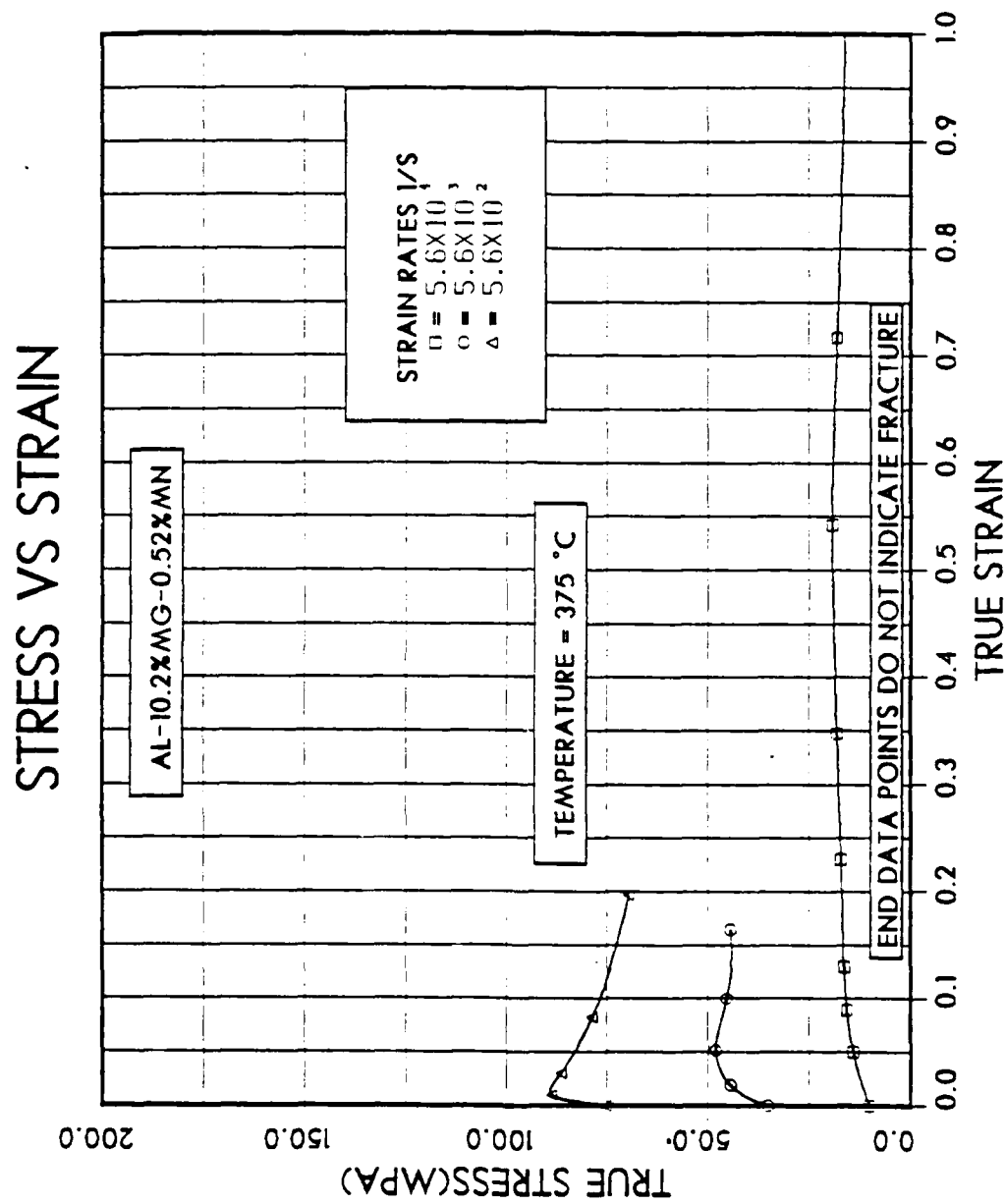


Figure A.10 True Stress vs True Plastic Strain Data for Testing Conducted at 375 C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 Hours, Annealed at 440 C for 1 Hour, Oil quenched, and Warm Rolled at 300 C to 94% Reduction.

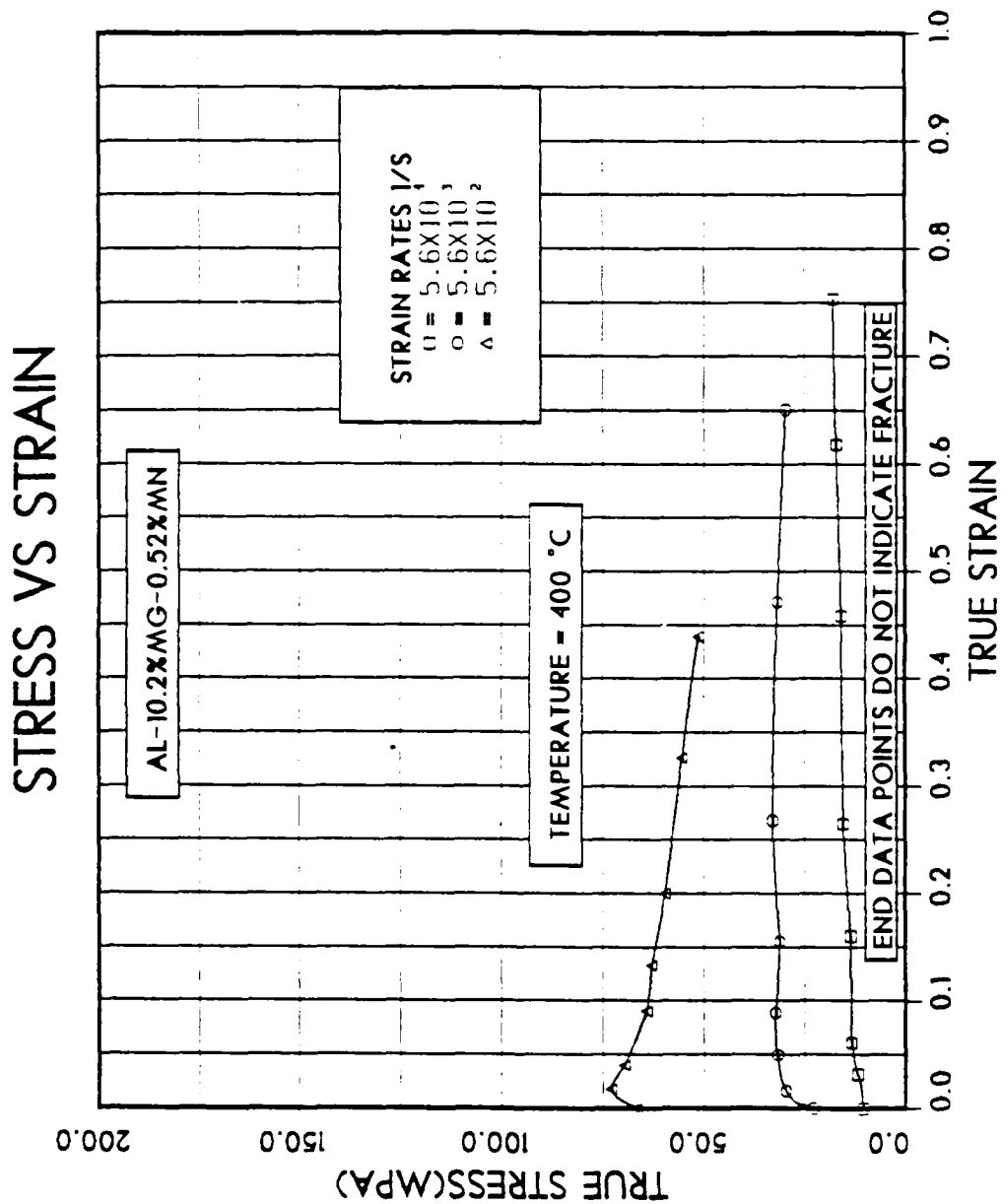


Figure A.11 True Stress vs True Plastic Strain Data for Testing Conducted at 400 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 hour, oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

STRESS VS STRAIN

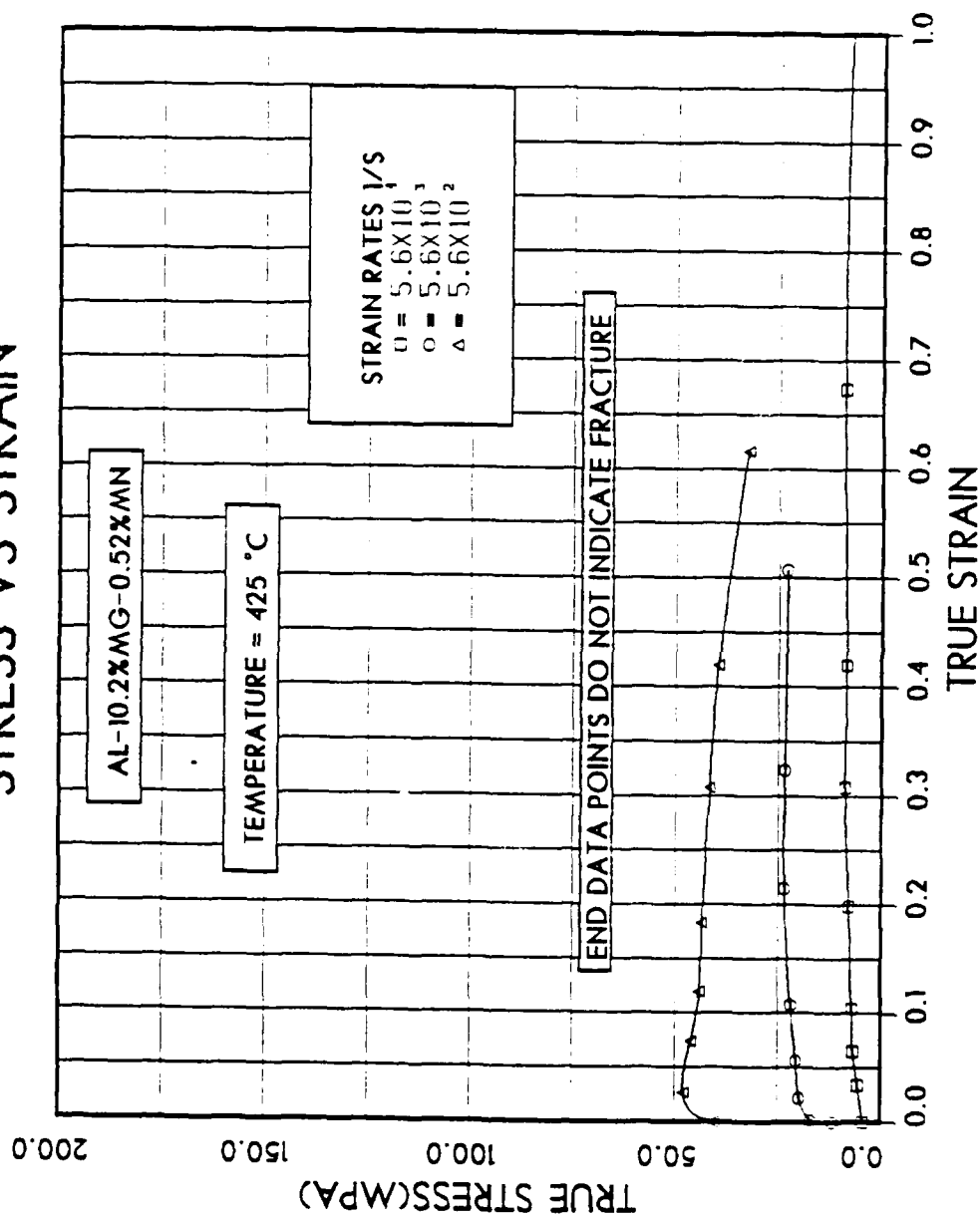


Figure A.12 True Stress vs True Plastic Strain Data for Testing Conducted at 425 °C for Al-10.2Mg-0.52Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

DUCTILITY VS TEMPERATURE

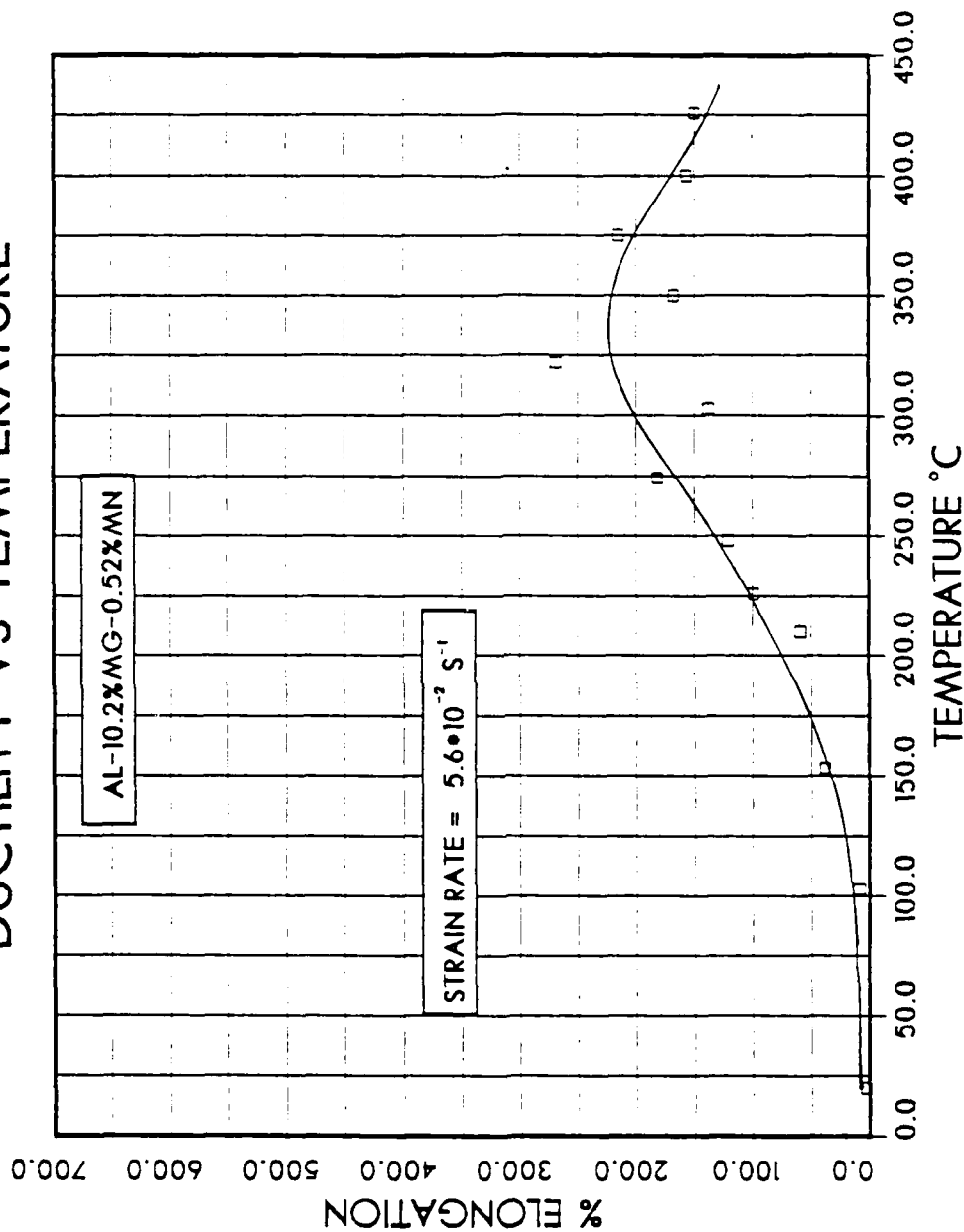


Figure A.13 Ductility vs Temperature Data Comparing As-Rolled to Recrystallized Data for Testing Conducted at $5.6 \times 10^{-2} \text{ S}^{-1}$ for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 C for 24 Hours, Annealed at 440 C for 1 Hour, Oil Quenched, and Warm Rolled at 300 C to 94% Reduction.

DUCTILITY VS TEMPERATURE

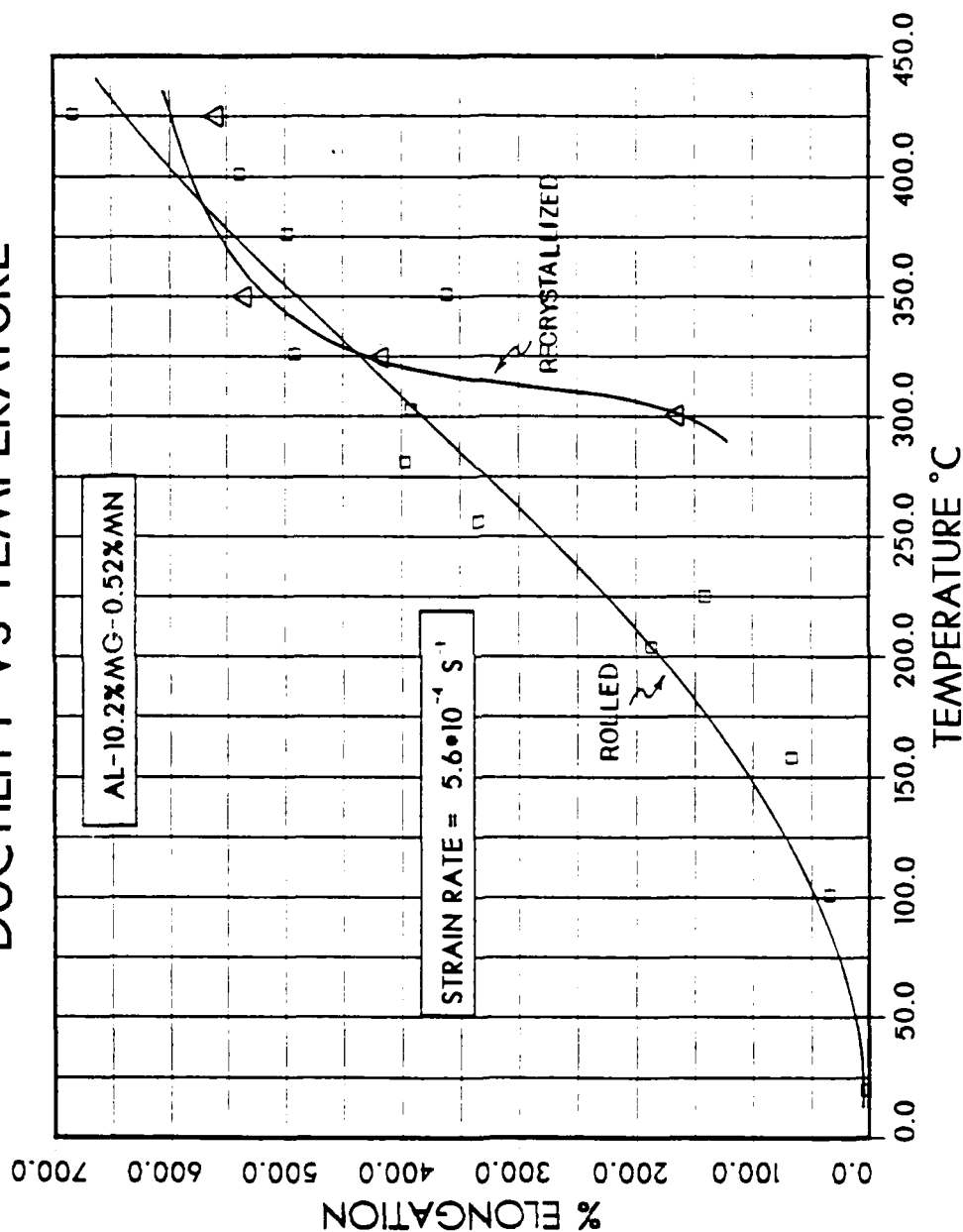


Figure A.14 Ductility vs Temperature Data Comparing As-Rolled to Recrystallized Data for Testing Conducted at $5.6 \times 10^{-4} \text{ s}^{-1}$ for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

DUCTILITY VS STRAIN RATE

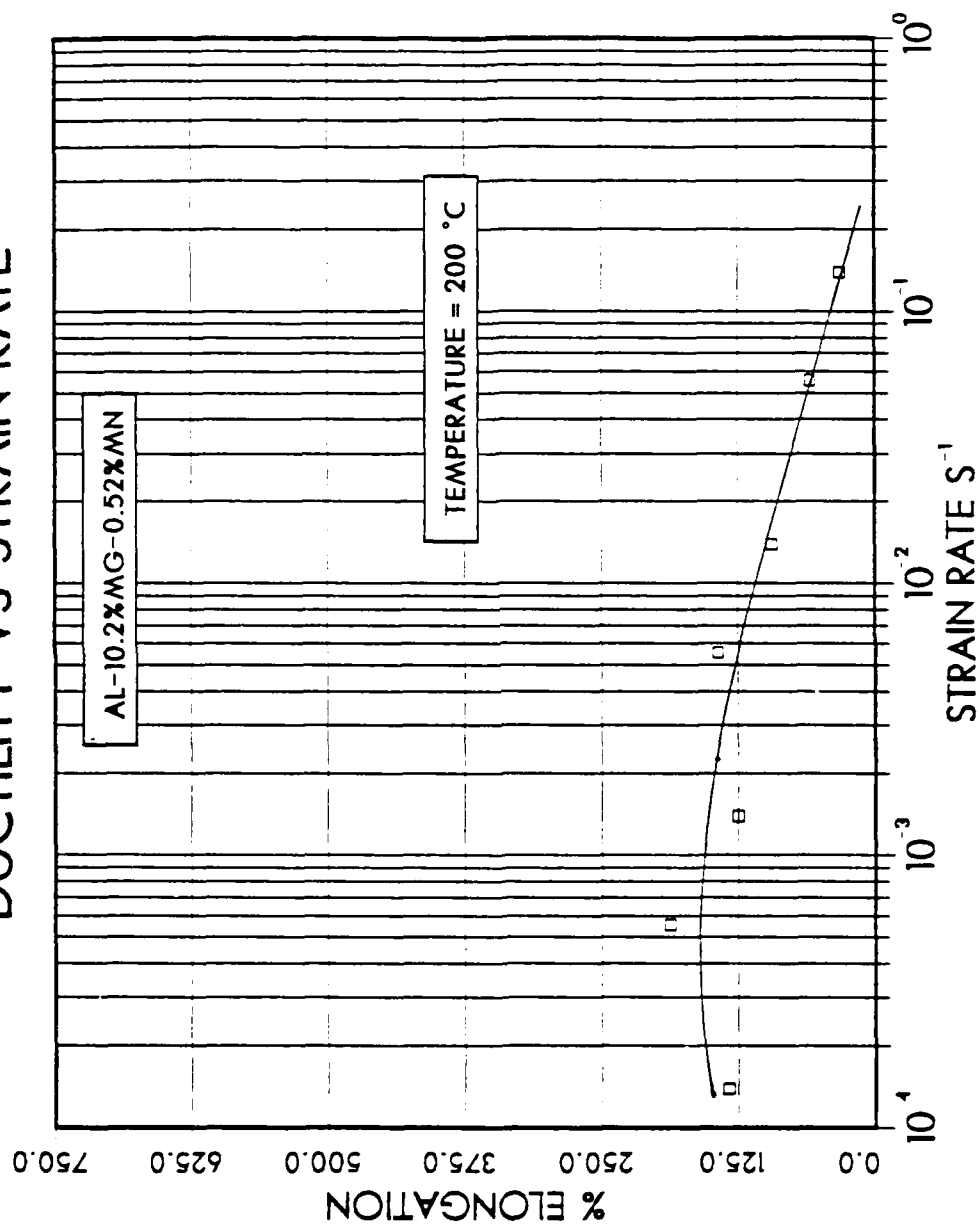


Figure A.15 Ductility vs Strain Rate Data for Testing Conducted at 200 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

DUCTILITY VS STRAIN RATE

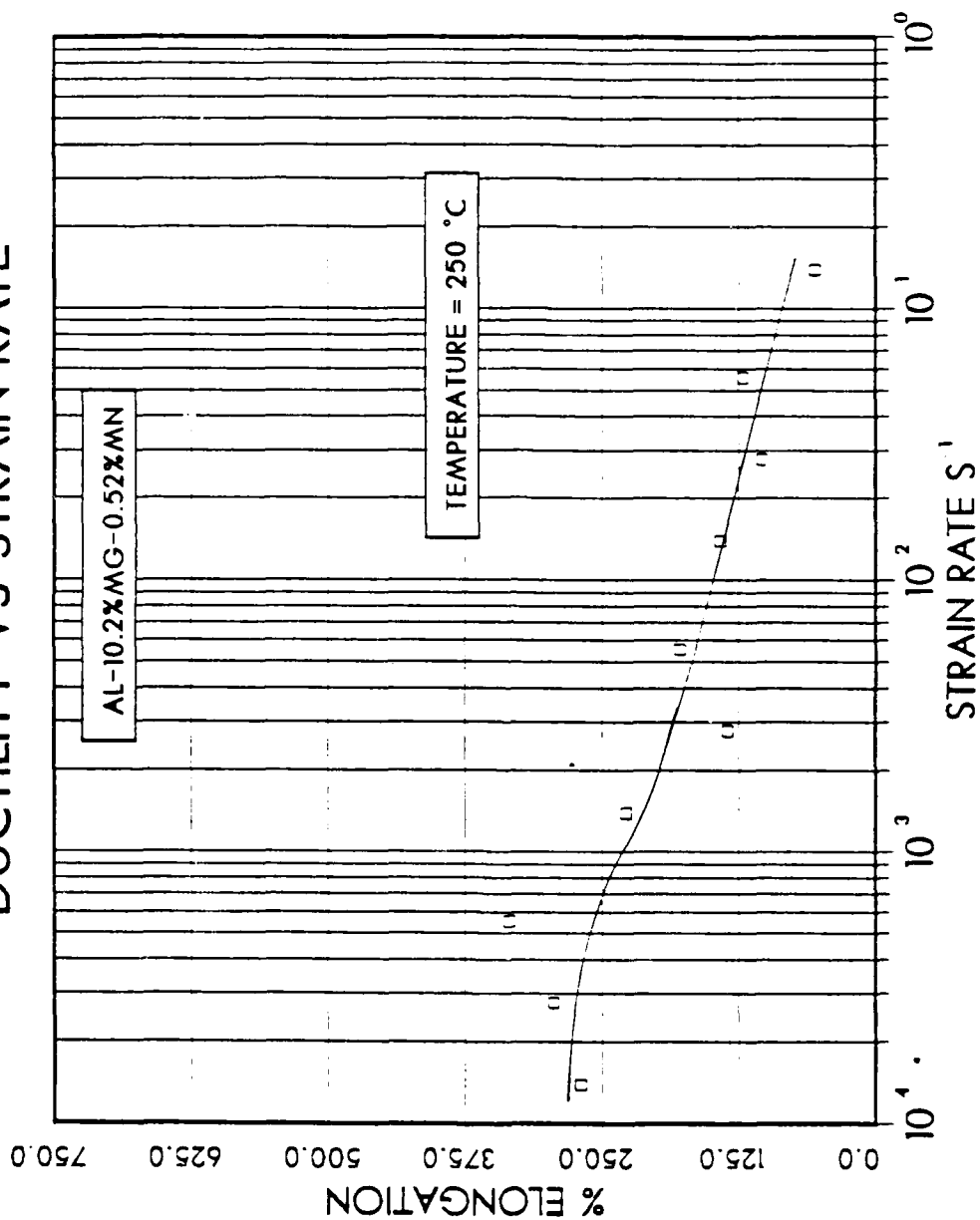


Figure A.16 Ductility vs Strain Rate Data for Testing Conducted at 250 °C for AL-10.2%MG-0.52%Mn. Solution Treated at 440 °C for 24 hours, Annealed at 440 °C for 1 hour, oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

DUCTILITY VS STRAIN RATE

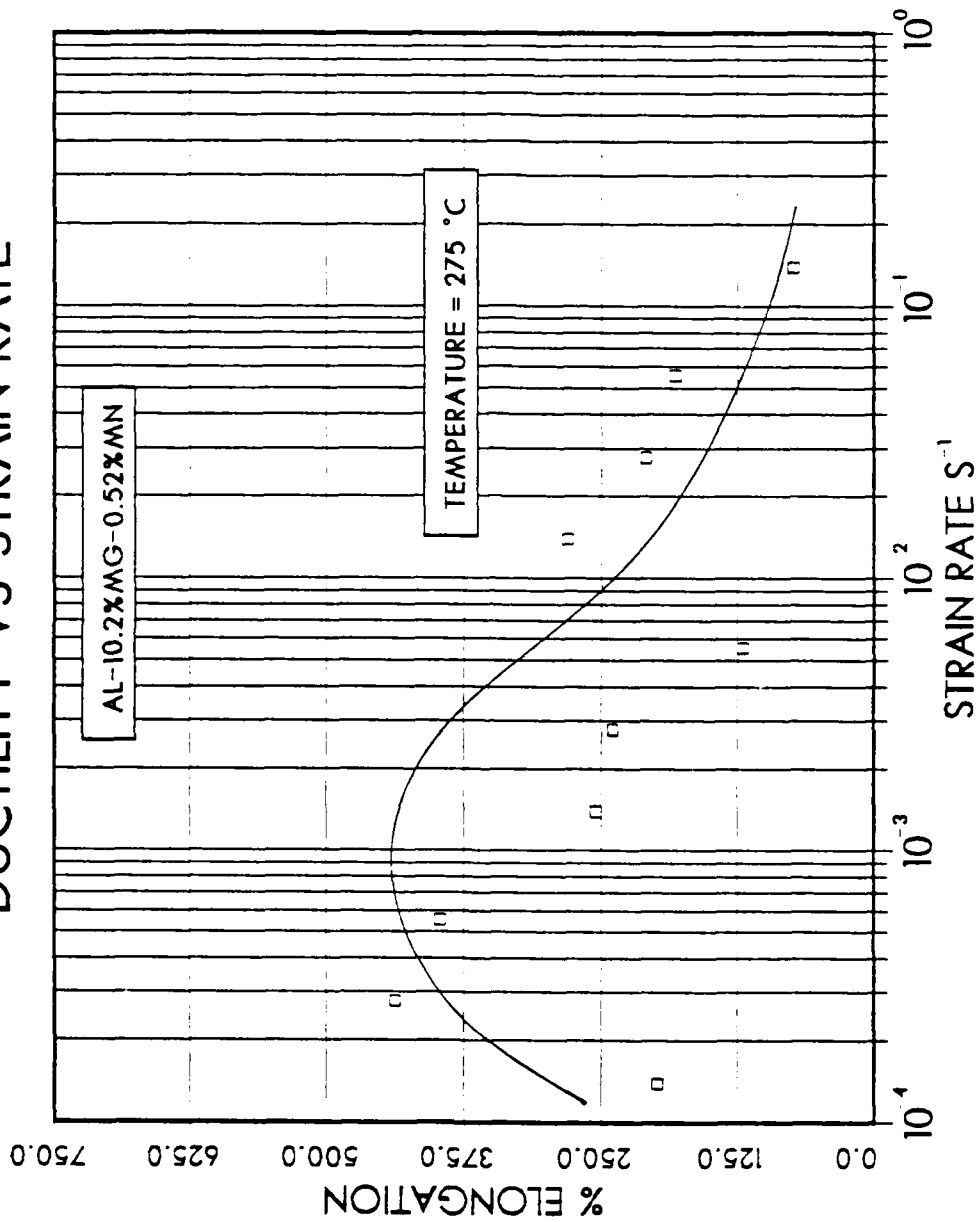


Figure A.17 Ductility vs Strain Rate Data for Testing Conducted at 275 °C for Al-10.2%Mg-0.52 Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

DUCTILITY VS STRAIN RATE

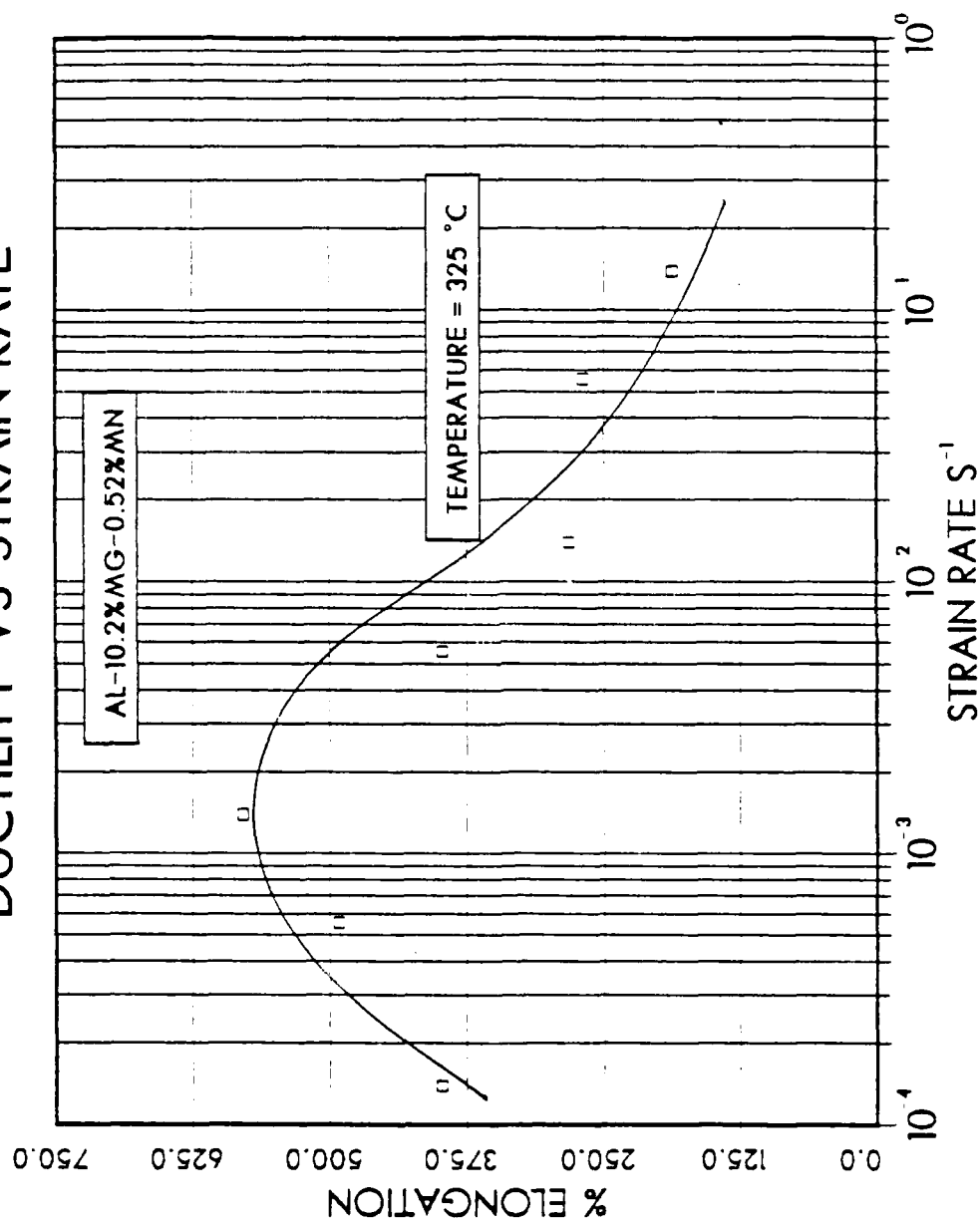


Figure A.18 Ductility vs Strain Rate Data for Testing Conducted at 325 °C for Al-10.2%Mg-0.52%Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil quenched, and Warm Rolled at 300 °C to 94% Reduction.

DUCTILITY VS STRAIN RATE

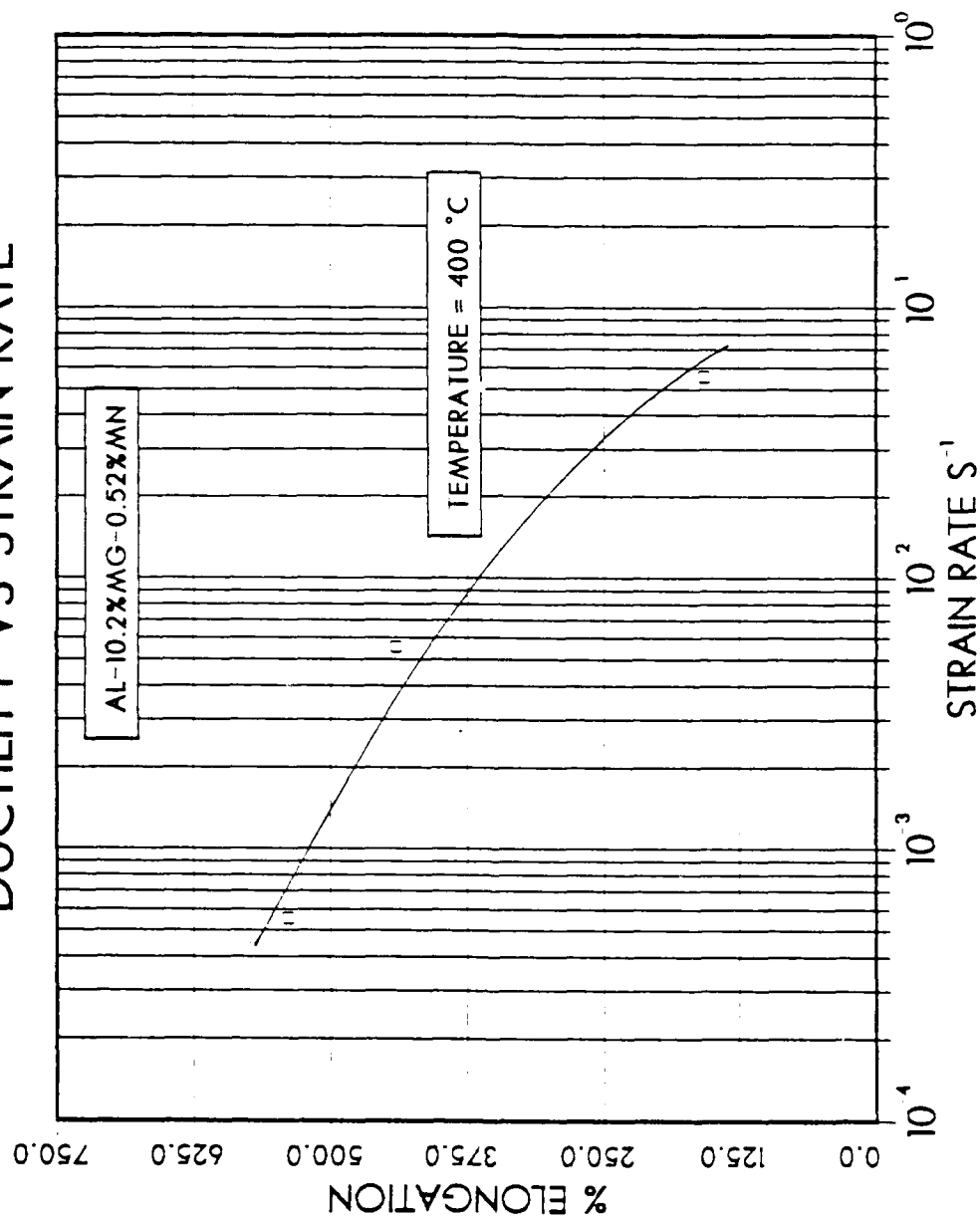


Figure A.19 Ductility vs Strain Rate Data for Testing Conducted at 400 °C for Al-10.2%Mg-0.52% Mn. Solution Treated at 440 °C for 24 Hours, Annealed at 440 °C for 1 Hour, Oil Quenched, and Warm Rolled at 300 °C to 94% Reduction.

APPENDIX B

```

C      TRUE STRESS VS TRUE STRAIN = 3000
C      THIS PROGRAM CALCULATES TRUE STRESS AND STRAIN FROM INPUT FILE OF
C      ENGINEERING STRESS AND STRAIN, AND THEN PLOTS TRUE
C      STRESS AGAINST TRUE STRAIN.
C      *****
C      EXTERNAL SLOPE
C      REAL A1(10),A2(10),A3(10),A4(10),A5(10),A6(10)
C      REAL A7(10),A8(10),A9(10),A10(10),LECPAK(500)
C      REAL B1(10),B2(10),B3(10),B4(10),B5(10),B6(10)
C      REAL S1(10),S2(10),S3(10),S4(10),S5(10),S6(10)
C      REAL E1(10),E2(10),E3(10),E4(10),E5(10),E6(10)
C      REAL C,C,E,S,CHG,TC,TO
C      INTEGER PTS1,PTS2,PTS3,PTS4,PTS5,PTS6
C      INTEGER PTS7,PTS8,PTS9,PTS10
C      I=0
C      WRITE(6,5)
C      CONTINUE
C      I=I+1
C      READ(57,*,END=20)A1(I),B1(I)
C      S1(I)=A1(I)*(1+B1(I))
C      E1(I)=ALCG(B1(I)+1)
C      ADJUSTMENT FOR INSTRON *****
C      C=5.34
C      D=.05
C      E=E1(I)
C      S=S1(I)
C      CALL SLOPE(C,C,E,S,CHG)
C      E1(I)=CHG
C      *****
C      WRITE(6,1)A1(I),S1(I),B1(I),E1(I)
C      GC TO 10
C      CONTINUE
C      PTS1=I-1
C      11111111111111111111111111111111
C      I=0
C      WRITE(6,5)
C      CONTINUE
C      I=I+1
C      READ(54,*,END=40)A2(I),B2(I)
C      S2(I)=A2(I)*(1+B2(I))
C      E2(I)=ALCG(B2(I)+1)
C      ADJUSTMENT FOR INSTRON *****
C      C=10.4
C      D=.1
C      E=E2(I)
C      S=S2(I)
C      CALL SLOPE(C,C,E,S,CHG)
C      E2(I)=CHG
C      *****
C      WRITE(6,2)A2(I),S2(I),B2(I),E2(I)
C      GC TO 30
C      CONTINUE
C      PTS2=I-1
C      22222222222222222222222222222222
C      I=0
C      WRITE(6,5)
C      CONTINUE
C      I=I+1
C      READ(56,*,END=60)A3(I),B3(I)
C      S3(I)=A3(I)*(1+B3(I))
C      E3(I)=ALCG(B3(I)+1)
C      ADJUSTMENT FOR INSTRON *****
C      C=11.9
C      D=.017
C      E=E3(I)
C      S=S3(I)
C      CALL SLOPE(C,C,E,S,CHG)
C      E3(I)=CHG
C      *****
C      WRITE(6,3)A3(I),S3(I),B3(I),E3(I)
C      GC TO 50
C      CONTINUE
C      PTS3=I-1
C      33333333333333333333333333333333
C      I=0
C      WRITE(6,5)
C      CONTINUE
C      I=I+1
C      READ(55,*,END=80)A4(I),B4(I)
C      S4(I)=A4(I)*(1+B4(I))
C      E4(I)=ALCG(B4(I)+1)
C      ADJUSTMENT FOR INSTRON *****
C      C=14.2
C      D=.05
C      E=E4(I)
C      S=S4(I)
C      CALL SLOPE(C,C,E,S,CHG)
C      E4(I)=CHG
C      *****

```



```

      CALL MESSAGE('END DATA POINTS CC NOT 5',100,1,2,1)
      CALL MESSAGE('INDICATE FRACTURES',100,'ABUT','ABUT')
      CALL BLREC(1,1,0.06,0.09,0.24,0.02)
      CALL JRTD(2,2)
      CALL ENDFL(0)
      CALL JCNEFL
      FORMAT(1X,4F12.5)
      FORMAT(1X,4F12.5)
      FORMAT(1X,4F12.5)
      FORMAT(1X,13)
      FORMAT(1X,13)
      FORMAT(1X,/,4X,'ENG STRESS',2X,'TRUE STRESS',2X,'ENG STRAIN',2X,
9 'TRUE STRAIN',/)
      STOP
      END
      SUBROUTINE SLCPE(C,C,E,S,CHG)
      REAL C,C,E,S,CHG,TC,TD
      TC=C*(1.+0)
      TC=ALOG(C+1.)
      CHG=E-S*TD/TC
      IF(CHG.LE.0.)GC TO 11
      GC TO 21
      CHG=0.
      CONTINUE
      RETURN
      END
11
21

```


74

```

      IF(I.GT.58)GC TO 8
      J=I-1
      C      WRITE(6,2000)J
      S325(J)=A(I)*1.36
      SR325(J)=B(I)*.001
      C      WRITE(6,1000)S325(J),SR325(J)
      GC TO 2000
      C      CONTINUE
      IF(I.GT.59)GC TO 9
      J=I-1
      S350(J)=A(I)*1.36
      SR350(J)=B(I)*.001
      C      WRITE(6,2000)J
      C      WRITE(6,1000)S350(J),SR350(J)
      GC TO 2000
      C      CONTINUE
      C      =====
      IF(I.GT.62)GC TO 10
      J=I-1
      C      WRITE(6,2000)J
      S375(J)=A(I)*1.36
      SR375(J)=B(I)*.001
      C      WRITE(6,1000)S375(J),SR375(J)
      GC TO 2000
      C      CONTINUE
      IF(I.GT.65)GC TO 11
      J=I-1
      S400(J)=A(I)*1.36
      SR400(J)=B(I)*.001
      C      WRITE(6,2000)J
      C      WRITE(6,1000)S400(J),SR400(J)
      GC TO 2000
      C      CONTINUE
      C      =====
      J=I-1
      C      WRITE(6,2000)J
      S425(J)=A(I)*1.36
      SR425(J)=B(I)*.001
      C      WRITE(6,1000)S425(J),SR425(J)
      GC TO 2000
      C      CONTINUE
      100      CALL COMPARS
      CALL BLCKOUT(.85)
      C      CALL SMCCCH
      C      CALL PSMTH
      CALL PAGE(1,1,2,5)
      MAXLINE=LINEENT(1,CPAK,3,20)
      C      CALL LINEOUT(1,CPAK,1)
      C      CALL LINEOUT(1,CPAK,2)
      C      CALL LINEOUT(1,CPAK,3)
      C      CALL LINEOUT(1,CPAK,4)
      C      CALL LINEOUT(1,CPAK,5)
      C      CALL LINEOUT(1,CPAK,6)
      C      CALL LINEOUT(1,CPAK,7)
      C      CALL LINEOUT(1,CPAK,8)
      C      CALL LINEOUT(1,CPAK,9)
      C      CALL LINEOUT(1,CPAK,10)
      C      CALL LINEOUT(1,CPAK,11)
      C      CALL LINEOUT(1,CPAK,12)
      C      CALL LINEOUT(1,CPAK,13)
      C      CALL LINEOUT(1,CPAK,14)
      C      CALL LINEOUT(1,CPAK,15)
      C      CALL LINEOUT(1,CPAK,16)
      C      CALL LINEOUT(1,CPAK,17)
      C      CALL LINEOUT(1,CPAK,18)
      C      CALL LINEOUT(1,CPAK,19)
      C      CALL LINEOUT(1,CPAK,20)

```



```

C      PLT5 = ELONG / 5. LOG STRAIN / 415.
C      *****
C      DIMENSION A(100),B(100),S20(10),SR20(10),S100(10),SR100(10)
C      DIMENSION S150(10),SR150(10),S200(10)
C      DIMENSION SR200(10),S225(10),SR225(10),S250(10),SR250(10)
C      DIMENSION S275(10),SR275(10),S300(10),SR300(10),S325(10)
C      DIMENSION SR325(10),S350(10),SR350(10),S375(10),SR375(10)
C      DIMENSION S400(10),SR400(10),S425(10),SR425(10),C(100)
C      INTEGER I,J
C      INTEGER PTS1,PTS2,PTS3,PTS4,PTS5,PTS6,PTS7,PTS8,PTS9
C      INTEGER PTS10,PTS11,PTS12,PTS13
C      PTS1=3
C      PTS2=3
C      PTS3=4
C      PTS4=7
C      PTS5=7
C      PTS6=10
C      PTS7=10
C      PTS8=10
C      PTS9=7
C      PTS10=3
C      PTS11=3
C      PTS12=3
C      PTS13=3
C      I=0
200  CONTINUE
C      I=I+1
C      *WRITE(6,2000)I
C      READ(96,*,END=100)A(I),B(I),C(I)
C      *WRITE(6,3000)A(I),B(I),C(I)
C      IF(I.GT.3)GO TO 1
C      S20(I)=A(I)
C      SR20(I)=B(I)*.001
C      *WRITE(6,1000)S20(I),SR20(I)
C      GO TO 200
1  CONTINUE
C      IF(I.GT.6)GC TC 2
C      J=I-3
C      *WRITE(6,2000)J
C      S100(J)=A(I)
C      SR100(J)=B(I)*.001
C      *WRITE(6,1000)S100(J),SR100(J)
C      GO TO 200
2  CONTINUE
C      IF(I.GT.10)GC TC 3
C      J=I-6
C      *WRITE(6,2000)J
C      S150(J)=A(I)
C      SR150(J)=B(I)*.001
C      *WRITE(6,1000)S150(J),SR150(J)
C      GO TO 200
3  CONTINUE
C      *****
C      IF(I.GT.17)GC TC 4
C      J=I-10
C      *WRITE(6,2000)J
C      S200(J)=A(I)
C      SR200(J)=B(I)*.001
C      *WRITE(6,1000)S200(J),SR200(J)
C      GO TO 200
4  CONTINUE
C      IF(I.GT.24)GC TC 5
C      J=I-17
C      S225(J)=A(I)
C      SR225(J)=B(I)*.001
C      *WRITE(6,2000)J
C      *WRITE(6,1000)S225(J),SR225(J)
C      GO TO 200
5  CONTINUE
C      *****
C      IF(I.GT.34)GC TC 6
C      J=I-24

```



```

CALL THKCRV(42,1,1,100,1)
CALL THKCRV(42)
CALL HEIGHT(1,2)
CALL XNAME('STRAIN RATE S(EH.3)-1E*.100)
CALL YNAME('X = ELONGATION*.100)
CALL AREA2D(2,6)
CALL HEACIN(' 1*.100,5.2)
CALL HEACIN('DUCTILITY VS STRAIN RATE*.100,1,5.2)
CALL XLOG(.0001,2,0,125)
CALL THKFRM(102)
CALL FRAME
C CALL CURVE(SR20,S20,PTS1,-1)
C CALL CURVE(SR100,S100,PTS2,-1)
C CALL CURVE(SR150,S150,PTS3,-1)
C CALL CURVE(SR200,S200,PTS4,-1)
C CALL CURVE(SR225,S225,PTS5,-1)
C CALL CURVE(SR250,S250,PTS6,-1)
C CALL CURVE(SR275,S275,PTS7,-1)
C CALL CURVE(SR300,S300,PTS8,-1)
C CALL CURVE(SR325,S325,PTS9,-1)
C CALL CURVE(SR350,S350,PTS10,-1)
C CALL CURVE(SR375,S375,PTS11,-1)
C CALL CURVE(SR400,S400,PTS12,-1)
C CALL CURVE(SR425,S425,PTS13,-1)
CALL RESET('THKCRV')
CALL RESET('HEIGHT')
C *****
CALL MESSAGE('TEMPERATURE = 5*.100,4,5,3)
CALL INTNO(425,'ABUT','ABUT')
CALL MESSAGE(' (EH.3)C(EXHX)CS*.100,'ABUT','ABUT')
CALL BLREC(4,3,2,9,2,7,4,02)
C *****
C *****
CALL MESSAGE('AL-10.2XMG-0.52XMN1*.100,3,5,5)
CALL BLREC(2,8,5,4,2,6,4,02)
C *****
CALL GRID(1,1)
CALL ENDPL(0)
CALL DONEPL
1000 FORMAT(1X,F20.5,1X,F20.12)
2000 FORMAT(1X,13)
3000 FORMAT(1X,JF12.5)
STOP
END

```



```

C      TRUE STRESS VS TRUE STRAIN AT T=1000°C
C      THIS PROGRAM READS TRUE STRESS AND STRAIN FROM INPUT FILES OF
C      ENGINEERING STRESS AND STRAIN, AND THEN PLOTS TRUE
C      STRESS AGAINST TRUE STRAIN.
C      *****
EXTERNAL SLCPF
REAL A1(10),A2(10),A3(10),B1(10),B2(10),B3(10)
REAL S1(10),S2(10),S3(10),E1(10),E2(10),E3(10)
REAL C,C,E,S,CHG,TC,TD,LEGPAK(500)
INTEGER I,PTS1,PTS2,PTS3
I=0
WRITE(6,5)
CONTINUE
10  I=I+1
    READ(81,*,END=20)A1(I),B1(I)
    S1(I)=A1(I)*(1+B1(I))
    E1(I)=ALCG(B1(I)+1)
C    ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
    C=350.
    D=.1
    S=S1(I)
    E=E1(I)
    CALL SLCPF(C,C,E,S,CHG)
    E1(I)=CHG
C    *****
    WRITE(6,1)A1(I),S1(I),B1(I),E1(I)
    GO TO 10
CONTINUE
20  PTS1=I-1
    I=0
    WRITE(6,5)
    CONTINUE
30  I=I+1
    READ(81,*,END=40)A2(I),B2(I)
    S2(I)=A2(I)*(1+B2(I))
    E2(I)=ALCG(B2(I)+1)
C    ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
    C=12.
    D=.033
    S=S2(I)
    E=E2(I)
    CALL SLCPF(C,C,E,S,CHG)
    E2(I)=CHG
C    *****
    WRITE(6,2)A2(I),S2(I),B2(I),E2(I)
    GO TO 30
CONTINUE
40  PTS2=I-1
    I=0
    WRITE(6,5)
    CONTINUE
50  I=I+1
    READ(80,*,END=60)A3(I),B3(I)
    S3(I)=A3(I)*(1+B3(I))
    E3(I)=ALCG(B3(I)+1)
C    ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
    C=436.
    D=.13
    S=S3(I)
    E=E3(I)
    CALL SLCPF(C,C,E,S,CHG)
    E3(I)=CHG
C    *****
    WRITE(6,2)A3(I),S3(I),B3(I),E3(I)
    GO TO 50
CONTINUE
60  PTS3=I-1
C    ***** 11111 DIMENSION LEGPAK 11111 *****
    CALL CCMFES
    CALL SCLY2
    CALL SLCWCP(.E.E)
    CALL PAGE(11.,E.E)

```



```

C      TRUE STRAIN VS TRUE STRAIN AT T=1500
C      THIS PROGRAM CONVERTS TRUE STRAIN FROM INPUT FILE
C      ENGINEERING STRESS AND STRAIN, AND THEN PLOTS TRUE
C      STRESS AGAINST TRUE STRAIN.
C      *****
C      EXTERNAL SLOPE
C      REAL A1(10),A2(10),A3(10),B1(10),B2(10),B3(10)
C      REAL S1(10),S2(10),S3(10),E1(10),E2(10),E3(10)
C      REAL A4(10),A5(10),A6(10),B4(10),B5(10),B6(10)
C      REAL C,D,E,S,CHG,TC,TD
C      INTEGER I,PTS1,PTS2,PTS3,PTS4
C      I=0
C      WRITE(6,5)
10    CCNTINUE
C      I=I+1
C      READ(80,*,END=20)A1(I),B1(I)
C      S1(I)=A1(I)*(1+B1(I))
C      E1(I)=ALCG(B1(I)+1)
C      ACJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=218.
C      D=.13
C      S=S1(I)
C      E=E1(I)
C      CALL SLCPE(C,D,E,S,CHG)
C      E1(I)=CHG
C      *****
C      WRITE(6,1)A1(I),S1(I),B1(I),E1(I)
C      GC TO 10
20    CCNTINUE
C      PTS1=I-1
C      I=0
C      WRITE(6,5)
30    CCNTINUE
C      I=I+1
C      READ(47,*,END=40)A2(I),B2(I)
C      S2(I)=A2(I)*(1+B2(I))
C      E2(I)=ALCG(B2(I)+1)
C      ACJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=255.
C      D=.1
C      S=S2(I)
C      E=E2(I)
C      CALL SLCPE(C,D,E,S,CHG)
C      E2(I)=CHG
C      *****
C      WRITE(6,2)A2(I),S2(I),B2(I),E2(I)
C      GC TO 30
40    CCNTINUE
C      PTS2=I-1
C      I=0
C      WRITE(6,5)
50    CCNTINUE
C      I=I+1
C      READ(46,*,END=60)A3(I),B3(I)
C      S3(I)=A3(I)*(1+B3(I))
C      E3(I)=ALCG(B3(I)+1)
C      ACJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=237.
C      D=.1
C      S=S3(I)
C      E=E3(I)
C      CALL SLCPE(C,D,E,S,CHG)
C      E3(I)=CHG
C      *****
C      WRITE(6,3)A3(I),S3(I),B3(I),E3(I)
C      GC TO 50
60    CCNTINUE
C      PTS3=I-1
C      I=0
C      WRITE(6,5)
70    CCNTINUE
C      I=I+1
C      READ(32,*,END=80)A4(I),B4(I)
C      S4(I)=A4(I)*(1+B4(I))
C      E4(I)=ALCG(B4(I)+1)
C      ACJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=334.
C      D=.1
C      S=S4(I)
C      E=E4(I)
C      CALL SLCPE(C,D,E,S,CHG)
C      E4(I)=CHG
C      *****
C      WRITE(6,3)A4(I),S4(I),B4(I),E4(I)
C      GC TO 70
80    CCNTINUE
C      PTS4=I-1
C      ***** 17227 DIMENSION LEGPAK 111111

```



```

C      TRUE STRAINS VS TRUE STRAIN AT T=200
C      TIME DEPENDENT STRESS AND STRAIN FROM INPUT FILE OF
C      STRESS AND STRAIN, AND THEN PLOTS TRUE
C      STRESS AND STRAIN.
C      *****
EXTERNAL SLCPE
REAL A1(10),A2(10),A3(10),A4(10),A5(10),A6(10)
REAL A7(10),A8(10),A9(10),A10(10),A11(10)
REAL B1(10),B2(10),B3(10),B4(10),B5(10),B6(10)
REAL B7(10),B8(10),B9(10),B10(10),B11(10)
REAL S1(10),S2(10),S3(10),S4(10),S5(10),S6(10)
REAL S7(10),S8(10),S9(10),S10(10),S11(10)
REAL E1(10),E2(10),E3(10),E4(10),E5(10),E6(10)
REAL E7(10),E8(10),E9(10),E10(10),E11(10)
REAL C,C2,C3,CHG,TC,T0,LEGPAC(500)
INTEGER I,PTS1,PTS2,PTS3,PTS4,PTS5,PTS6
INTEGER PTS7,PTS8,PTS9,PTS10
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(74,*,END=20)A1(I),B1(I)
S1(I)=A1(I)*(1+B1(I))
E1(I)=ALCG(B1(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=45.1
D=.333
S=S1(I)
E=E1(I)
CALL SLCPE(C,C,E,S,CHG)
E1(I)=CHG
*****
WRITE(6,1)A1(I),S1(I),B1(I),E1(I)
GC TO 10
CONTINUE
PTS1=I-1
C      11111111111111111111
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(77,*,END=40)A2(I),B2(I)
S2(I)=A2(I)*(1+B2(I))
E2(I)=ALCG(B2(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=105.
D=.15
S=S2(I)
E=E2(I)
CALL SLCPE(C,C,E,S,CHG)
E2(I)=CHG
*****
WRITE(6,2)A2(I),S2(I),B2(I),E2(I)
GC TO 30
CONTINUE
PTS2=I-1
C      22222222222222222222
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(45,*,END=60)A3(I),B3(I)
S3(I)=A3(I)*(1+B3(I))
E3(I)=ALCG(B3(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=105.
D=.1
S=S3(I)
E=E3(I)
CALL SLCPE(C,C,E,S,CHG)
E3(I)=CHG
*****
WRITE(6,3)A3(I),S3(I),B3(I),E3(I)
GC TO 50
CONTINUE
PTS3=I-1
C      33333333333333333333
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(44,*,END=80)A4(I),B4(I)
S4(I)=A4(I)*(1+B4(I))
E4(I)=ALCG(B4(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=113.
D=.1
S=S4(I)
E=E4(I)
CALL SLCPE(C,C,E,S,CHG)
E4(I)=CHG
*****

```


AD-A152 016

SUPERPLASTICITY IN A THERMO-MECHANICALLY PROCESSED
ALUMINUM-10.2%MG-0.52%ZNN ALLOY(U) NAVAL POSTGRADUATE
SCHOOL MONTEREY CA M E MILLS SEP 84

2/2

UNCLASSIFIED

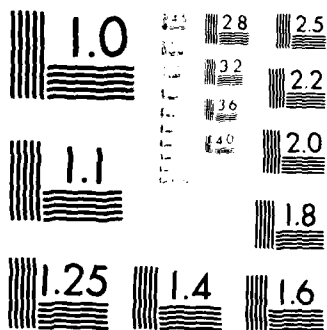
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MICROCOPY RESOLUTION TEST CHART
NATIONAL BUREAU OF STANDARDS-1963-A

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C      TRIP STRESS VS. STRAIN. AT T=35 IS
C      THE STRESS AT WHICH THE STRAIN FROM INPUT FILES
C      ENGINEERING STRESS AND STRAIN, AND THEN PLASTIC
C      STRESS AND STRAIN. *****
C      *****
EXTERNAL SLOPE
REAL A1(10),A2(10),A3(10),A4(10),A5(10),A6(10)
REAL A7(10),A8(10),A9(10),A10(10)
REAL B1(10),B2(10),B3(10),B4(10),B5(10),B6(10)
REAL B7(10),B8(10),B9(10),B10(10)
REAL S1(10),S2(10),S3(10),S4(10),S5(10),S6(10)
REAL S7(10),S8(10),S9(10),S10(10)
REAL E1(10),E2(10),E3(10),E4(10),E5(10),E6(10)
REAL E7(10),E8(10),E9(10),E10(10)
REAL C,C2,S,CHG,TC,TD,LEGPAC(500)
INTEGER PTS1,PTS2,PTS3,PTS4,PTS5,PTS6
INTEGER PTS7,PTS8,PTS9,PTS10
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(7,*,END=20)A1(I),B1(I)
S1(I)=A1(I)*(1+B1(I))
E1(I)=ALOG(B1(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=15.3
C=.375
S=S1(I)
E=E1(I)
CALL SLOPE(C,C,E,S,CHG)
E1(I)=CHG
C      *****
WRITE(6,1)A1(I),S1(I),B1(I),E1(I)
GO TO 10
CONTINUE
PTS1=-1
11111111111111111111
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(7,*,END=40)A2(I),B2(I)
S2(I)=A2(I)*(1+B2(I))
E2(I)=ALOG(B2(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=20.1
C=.342
S=S2(I)
E=E2(I)
CALL SLOPE(C,C,E,S,CHG)
E2(I)=CHG
C      *****
WRITE(6,2)A2(I),S2(I),B2(I),E2(I)
GO TO 30
CONTINUE
PTS2=-1
22222222222222222222
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(7,*,END=60)A3(I),B3(I)
S3(I)=A3(I)*(1+B3(I))
E3(I)=ALOG(B3(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=27.5
C=.383
S=S3(I)
E=E3(I)
CALL SLOPE(C,C,E,S,CHG)
E3(I)=CHG
C      *****
WRITE(6,3)A3(I),S3(I),B3(I),E3(I)
GO TO 50
CONTINUE
PTS3=-1
33333333333333333333
I=0
WRITE(6,5)
CONTINUE
I=I+1
READ(7,*,END=80)A4(I),B4(I)
S4(I)=A4(I)*(1+B4(I))
E4(I)=ALOG(B4(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C=37.0
C=.1
S=S4(I)
E=E4(I)
CALL SLOPE(C,C,E,S,CHG)
E4(I)=CHG
C      *****

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C      TRUE STRAIN VS TRUE STRESS AT T=4000
C      THIS SUBROUTINE CALCULATES TRUE STRESS AND STRAIN FROM INPUT FILES OF
C      ENGINEERING STRESS AND STRAIN, AND THEN PLOTS TRUE
C      STRESS AGAINST TRUE STRAIN.
C      *****
C      EXTERNAL SLCPF
C      REAL A1(10),A2(10),A3(10),B1(10),B2(10),B3(10)
C      REAL S1(10),S2(10),S3(10),E1(10),E2(10),E3(10)
C      REAL C,C2,S,CHG,TC,TD,LEGPAK(500)
C      INTEGER I,PTS1,PTS2,PTS3
C      I=0
10  WRITE(6,*)
C      CONTINUE
C      I=I+1
C      READ(90,*,END=20)A1(I),B1(I)
C      S1(I)=A1(I)*(1+B1(I))
C      E1(I)=ALOG(S1(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=9.51
C      D=.1
C      S=S1(I)
C      E=E1(I)
C      CALL SLCPF(C,C2,S,CHG)
C      E1(I)=C+E
C      *****
C      WRITE(6,1)A1(I),S1(I),B1(I),E1(I)
C      GC TO 10
20  CONTINUE
C      PTS1=I-1
C      1111111111111111111111111111111111
C      I=0
30  WRITE(6,*)
C      CONTINUE
C      I=I+1
C      READ(89,*,END=40)A2(I),B2(I)
C      S2(I)=A2(I)*(1+B2(I))
C      E2(I)=ALOG(S2(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=20.2
C      D=.1
C      S=S2(I)
C      E=E2(I)
C      CALL SLCPF(C,C2,S,CHG)
C      E2(I)=C+E
C      *****
C      WRITE(6,2)A2(I),S2(I),B2(I),E2(I)
C      GC TO 30
40  CONTINUE
C      PTS2=I-1
C      2222222222222222222222222222222222
C      I=0
50  WRITE(6,*)
C      CONTINUE
C      I=I+1
C      READ(88,*,END=60)A3(I),B3(I)
C      S3(I)=A3(I)*(1+B3(I))
C      E3(I)=ALOG(S3(I)+1)
C      ADJUSTMENT FOR INSTRON AND ELASTIC STRAIN *****
C      C=60.7
C      D=.1
C      S=S3(I)
C      E=E3(I)
C      CALL SLCPF(C,C2,S,CHG)
C      E3(I)=C+E
C      *****
C      WRITE(6,3)A3(I),S3(I),B3(I),E3(I)
C      GC TO 60
60  CONTINUE
C      PTS3=I-1
C      3333333333333333333333333333333333
C      *****
C      CALL COMPRS

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CALL SPACE(0,1)
CALL PAGE(11,0,0,0)
CALL MIXALR('INSTPL')
MAXLINE=LINEST(LSGPAK,500,20)
CALL LINEST('E*10(EH.5)-4',LSGPAK,1)
CALL LINEST('E*10(EH.5)-3',LSGPAK,2)
CALL LINEST('E*10(EH.5)-2',LSGPAK,3)
CALL MYLEGN('STRAIN RATES 1/S',16)
CALL PUTRA
CALL STECHQ(SC,1,002,1)
CALL THXGRV(.C2)
CALL HEIGHT(.C2)
CALL XNAME('TRUE STRAINS',100)
CALL YNAME('TRUE STRESS(IMPACT)',100)
CALL AREA2D(9,C,6,0)
CALL HEACIN(' ',100,5,2)
CALL HEACIN('STRESS VS STRAIN',100,1.5,2)
CALL GRAF(0,1,1,0,C,50,200)
CALL THXGRV(.C2)
CALL FRAM
CALL CURVE(1,1,1,FTS1,+1)
CALL CURVE(2,1,1,FTS2,+1)
CALL CURVE(3,1,1,FTS3,+1)
CALL CURVE(4,1,1,FTS4,+1)
CALL RESSET('TITKCFV')
CALL RESSET('TITKCFV')
CALL LEGEND(LSGPAK,3,5,4,3)
CALL BLREC(5,1,2,7,2,5,1.5,.C2)
C *****
CALL MESSAGE('TEMPERATURE = ',100,2,4,5)
CALL INTD(4,2,5,'ABUT','ABUT')
CALL MESSAGE(' (E*10(EH.5)-3)',100,'ABUT','ABUT')
CALL BLREC(1,8,4,2,7,4,.C2)
C *****
CALL MESSAGE('AL-10.2MG-C.521MNS',100,2,5,5)
CALL BLREC(2,2,5,4,2,0,4,.C2)
C *****
CALL MESSAGE('END DATA POINTS DO NOT ',100,1,2,2)
CALL MESSAGE('INDICATE FRACTURE',100,'ABUT','ABUT')
CALL BLREC(1,1,1,0,5,1,24,.C2)
C *****
CALL GRID(2,2)
CALL ENDPL(0)
CALL JCNEFL
1 FORMAT(1X,2F12.3,1X,2F12.7)
2 FORMAT(1X,4F12.3)
3 FORMAT(1X,4F12.3)
4 FORMAT(1X,3)
5 FORMAT(1X,7,4X,'TAG STRESS',2X,'TRUE STRESS',2X,'ENG STRAIN',2X,
a 'TRUE STRAIN',7)
STOP
END
C SUBROUTINE FOR CORRECTING INSTRON AND ELASTIC STRAIN *****
SUBROUTINE SLOPE(C,LS,S,CHG)
REAL C,C,LS,S,CHG,TC,TD
TC=C*(1.+0)
TC=ALOG(C+1)
CHG=LS-S*TD/TC
IF(CHG.LE.0)GO TO 11
GO TO 21
11 C=C*0.
21 CONTINUE
RETURN
END

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