THE 4TH INTERNATIONAL CONFERENCE ON ALUMINUM ALLOYS

WARM FORMABILITY OF ALUMINUM–MAGNESIUM ALLOYS

Eric M. Taleff [1], Gregory A. Henshall [2], Donald R. Lesuer [3], and T. G. Nieh [2]
1. Stanford University, Department of Mechanical Engineering, Stanford, CA 94305.
2. Lawrence Livermore National Laboratory, Chemistry and Materials Science Department, Livermore, CA 94550.
3. Lawrence Livermore National Laboratory, Manufacturing and Materials Engineering Division, Livermore, CA 94550.

1 Abstract

Manufacturers have become increasingly interested in near-net-shape forming of aluminum alloys as a means to reduce production costs and the weight of aircraft and automotive structures. To achieve the ductilities required for this process, we have examined extended ductility of Al–Mg alloys in the warm forming, or Class I creep, regime. We have studied a high-purity, binary alloy of Al–2.8Mg and ternary alloys of Al–xMg–0.5Mn with Mg concentrations from 1.0 to 6.6 wt.%. Tensile tests, including strain-rate-change tests, have been performed with these materials at temperatures of 300 and 400°C over a range in strain rates from 10^{-4} to 2×10^{-2} s^{-1}. A maximum tensile failure strain of 325% for the binary alloy and a maximum of 125% in the ternary alloys have been measured. The experimental results have been used to evaluate the effects of solute concentration, microstructure, temperature, and strain rate on the flow stress (σ), elongation to failure (e_f), and strain-rate sensitivity (m) of these alloys.

2 Introduction

Enhanced tensile ductilities, to over 300%, have been observed at warm temperatures in coarse-grained, solid solution alloys of aluminum containing magnesium [1, 2, 3]. This is

especially remarkable because none of these materials exhibit the grain-boundary sliding behavior that is typical of most fine-grained, superplastic metallic alloys. The ability of coarse-grained materials to exhibit high tensile ductilities is of great interest for economically producing materials with high formabilities, without the need for expensive processing procedures. High formability permits a cost-effective approach to the manufacturing of complex components through a reduction of the required stamping, machining, and joining operations.

The enhanced tensile ductilities observed in these coarse-grained Al–Mg alloys are not the result of a grain-boundary sliding deformation mechanism, but rather the result of a solute-drag controlled creep mechanism [4]. Alloys which exhibit such solute-drag controlled creep behavior, referred to here as Class I alloys after the original definition by Sherby and Burke [5], exhibit a high strain-rate-sensitivity value of $m \approx 0.33$. By contrast, pure aluminum and aluminum alloys with a low magnesium content do not exhibit high tensile ductilities at warm temperatures; these materials show a low strain-rate sensitivity of $m \approx 0.2$ [5]. Increasing values of m generally produce increases in ductility for metallic materials [6, 7].

As an augmentation of previous efforts, further systematic studies of the influence of Mg and ternary solute additions on the ductility and strain-rate sensitivity of Al–Mg alloys are of great interest. The ability to understand and to accurately predict under which conditions high tensile ductility is achievable will become possible only after such studies are made. The importance of both strain-rate sensitivity and Mn content on the ductility of Al–Mg alloys is studied in order to further this purpose. In particular, the effects of temperature and strain rate on ductility are examined for both ternary alloys of Al–xMg–0.5 wt.%Mn, where $1.0 \leq x \leq 6.6$ wt.%, and a high-purity binary alloy of Al–2.8 wt.%Mg.

3 Experimental Procedures

Aluminum alloy ingots containing five different magnesium concentrations, and approximately 0.5 wt.% Mn each, were obtained from Kaiser Aluminum Center for Technology. The magnesium concentrations were 1.02, 2.52, 4.05, 5.51, and 6.64 weight percent. The initial alloy ingots were scalped approximately 3 mm per side after casting, to dimensions of 25 mm × 152 mm × 127 mm. The ingots were then reheated to 430°C and hot rolled from 25 mm to 12.7 mm. The hot mill oxide was removed by etching, and the materials were then cold rolled from 12.7 mm to 4.6 mm. The samples were cleaned and annealed at 400°C for 1 hr. Tensile samples were subsequently machined for testing.

The high purity binary Al–2.8 wt.%Mg alloy was received as a 76.2-mm-diameter casting. A 76-mm-long section was cut from the casting and homogenized by annealing in air at 500°C for 24 hrs. The height of the cylindrical casting was reduced by upset forging at 350°C, followed by warm rolling at 300°C to a thickness of 15.5 mm. This plate was cold rolled to a thickness of 4.6 mm, annealed for 30 min. at 450°C, and then flattened by hot pressing at 450°C. Tensile samples were machined from this sheet for testing.

Two types of tensile tests were conducted. The first type was an elongation-to-failure test, in which tensile samples were pulled to failure at a constant true strain rate. Elongation-to-failure tests were performed at several temperatures and strain rates for each alloy. The second type was a strain-rate-change test. Strain-rate-change tests were performed at two temperatures for each material: 300 and 400°C. For these tests, each sample was pulled at a series of discrete constant true strain rates, ranging from 10^{-4} s^{-1} to a maximum value of 2×10^{-2} s^{-1}. An initial prestrain was introduced at the beginning of each test to ensure a stable structure, and strain rates were then altered from low values to high.

4 Results and Discussion

A plot of the logarithm of strain rate versus the logarithm of flow stress is given for the temperatures of 300 and 400°C in Fig. 1. Data for the binary alloy, Al–2.8Mg, and a ternary alloy of similar Mg content, Al–2.5Mg–0.5Mn, are given. The notable differences between the binary and the ternary alloy are: (1) a lower slope (higher m value) for the binary alloy and (2) a higher flow stress for the ternary alloy at strain rates of 10^{-3} s^{-1} and below.

Based upon experiments performed with binary Al–Mg alloys and a commercial 5182 alloy containing 4.05 wt.% Mg and 0.26 wt.% Mn, Ayres [2, 3] suggested than additions of Mn to Al–Mg alloys may enhance ductility. Our results appear to show the opposite effect. Fig. 2 shows that for a given Mg content and testing condition, the binary alloy exhibits a higher value of m. For example, for tests performed at 400°C and 10^{-4} s^{-1}, the strain-rate sensitivity of the binary 2.8 wt.% Mg alloy is 0.31, while the corresponding Al–Mg–0.5Mn alloy exhibits an m value of only about 0.21. Thus, the difference in m for the two alloys shown in Fig. 1 is not due to differences in Mg concentrations. As mentioned earlier, a higher failure elongation is expected to result for higher values of m [6]. The data in Fig. 3 show that this is indeed the case for the alloys tested in this study. For tests performed at 300°C and 10^{-3} s^{-1}, the binary alloy has a failure elongation of 132%, while the corresponding Al–Mg–0.5Mn alloy has an elongation to failure of only about 100%. For

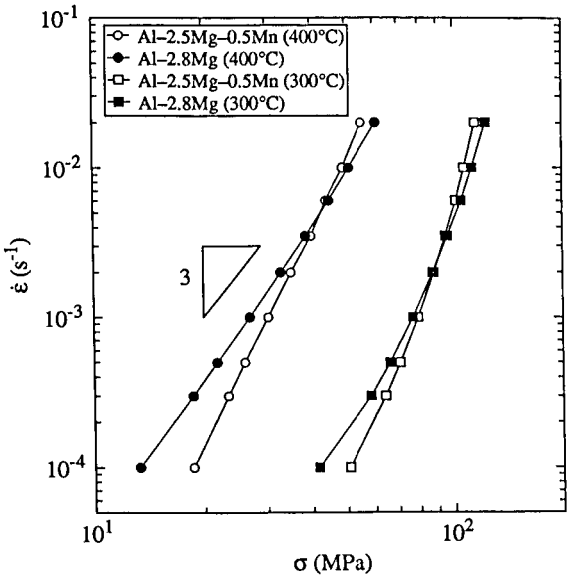

Figure 1: Logarithm of steady-state strain rate versus logarithm of flow stress for Al–2.8Mg and Al–2.5Mg–0.5Mn tested at 300 and 400°C.

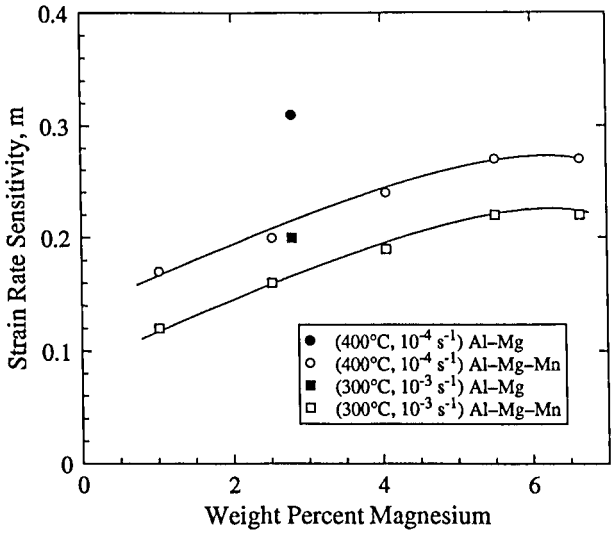

Figure 2: Strain-rate sensitivity as a function of Mg concentration for the binary Al–Mg alloy and the ternary Al–Mg–Mn alloys.

tests performed at 400°C and 10^{-4} s^{-1} the contrast is much more striking. The binary Al–Mg alloy exhibits an elongation to failure of 325% while the trend shown for the Al–Mg–0.5Mn alloys suggests a failure elongation of only about 100%. It is important to emphasize that the 325% failure elongation observed in the binary alloy is not the result of grain boundary sliding, since the grain size of this material is approximately 30 μm. Rather, the grain matrix deformation mechanism of viscous solute drag, with a corresponding m of 0.31, leads to the high tensile ductility of this alloy [4].

There are several possible reasons why the binary Al–Mg alloy exhibits better ductility than the Al–Mg–0.5Mn alloys. First, as discussed above, the addition of the Mn decreases m (Figs. 1 and 2). Second, for the low strain rate and high temperature test conditions resulting in the greatest ductility variation between the two alloys, the flow stress in the Mn-bearing material is higher than for the binary Al–Mg (Fig. 1). This higher flow stress could enhance cavitation, leading to decreased ductility for the ternary alloy. Finally, the Mn-bearing alloys contain second-phase particles in their microstructures [9]; the binary alloy was single phase. These second-phase particles, which have not yet been identified, could act as cavity nucleation sites and lower the failure elongation.

If the ductility of the binary Al–Mg alloys is controlled by the solute-drag deformation mechanism in the Class I regime, then the failure elongations for tests performed at different temperatures and strain rates should correlate with the diffusion-compensated strain rate, $\dot{\epsilon}/D$. As shown by McNelley et al. [8], the steady-state deformation of Al–Mg alloys tested under a variety of temperatures and strain rates can be unified using this variable, with D corresponding to diffusion of Mg in the matrix. To begin testing this hypothesis, the failure elongation is plotted as a function of $\dot{\epsilon}/D$ for the binary Al–Mg alloy in Fig. 4. Note the large increase in ductility as $\dot{\epsilon}/D$ decreases from values above 10^{14} to values below 10^{14} m^{-2}. As shown by McNelley et al. [8], the Class I deformation regime for Al–Mg alloys is limited to $\dot{\epsilon}/D < 10^{14}$ m^{-2}. Fig. 4 also shows that the highest elongation occurs at the lowest $\dot{\epsilon}/D$ tested, about 7×10^{10} m^{-2}, which is at the lower limit of Class I behavior. Since five-power creep ($m = 0.2$) behavior is exhibited for $\dot{\epsilon}/D$ below this value, the failure elongation may be expected to decrease for $\dot{\epsilon}/D < 7 \times 10^{10}$ m^{-2}. This prediction has yet to be tested, however. Perhaps the most interesting feature of Fig. 4 is related to the two data points near $\dot{\epsilon}/D = 5 \times 10^{13}$ m^{-2}. These two points represent the results of tests performed under widely different conditions: 300°C, 10^{-3} s^{-1} and 400°C, 10^{-1} s^{-1}. Since the $\dot{\epsilon}/D$ values are similar, however, the failure elongations should be similar if the solute-dislocation interactions control ductility. Fig. 4 shows that this is in fact the case. If the failure elongation were controlled by a cavity initiation and growth mechanism, for example, such a correlation between e_f and $\dot{\epsilon}/D$ might not occur.

Obviously, more data are required to determine if the correlation shown in Fig. 4 was fortuitous or if it truly represents the behavior of these alloys.

5 Conclusions

Strain-rate-change and elongation-to-failure experiments have been performed using a binary Al–2.8 wt.% Mg alloy and a range of Al–xMg–0.5Mn alloys. The results for these coarse-grained materials tested within or near the Class I, solute drag regime have led to the following conclusions:

1. The addition of Mn to Al–Mg alloys appears to decrease the ductility. The maximum failure elongation observed in the Al–Mg–0.5Mn alloys was 125%, whereas 325% elongation was achieved in the binary Al–Mg alloy. The decrease in ductility due to the presence of Mn can be explained, at least partially, by a decrease in strain-rate sensitivity. Other factors, such as an increase in flow stress and the presence of second-phase particles, also may contribute to the decrease in failure elongation for the Mn-bearing alloys.
2. Based on limited data, the tensile failure elongation of binary Al–Mg alloys appears to be controlled by the diffusion-compensated strain rate during testing. Tests performed under widely different temperature and strain rate conditions appear to have similar ductilities if the diffusion-compensated strain rates are similar. This finding suggests that the tensile ductility of binary Al–Mg alloys is controlled by the viscous drag of solute atmospheres by moving dislocations during grain matrix flow.

6 Acknowledgements

The authors would like to thank Mr. Rick Gross for running mechanical tests. One author, E.M.T., would like to express his gratitude to Lawrence Livermore National Laboratories for their assistance in the advancement of his Ph.D. thesis program at Stanford University. This work was performed under the auspices of the United States Department of Energy by the Lawrence Livermore National Laboratory under Contract No. W-7405-ENG-48.

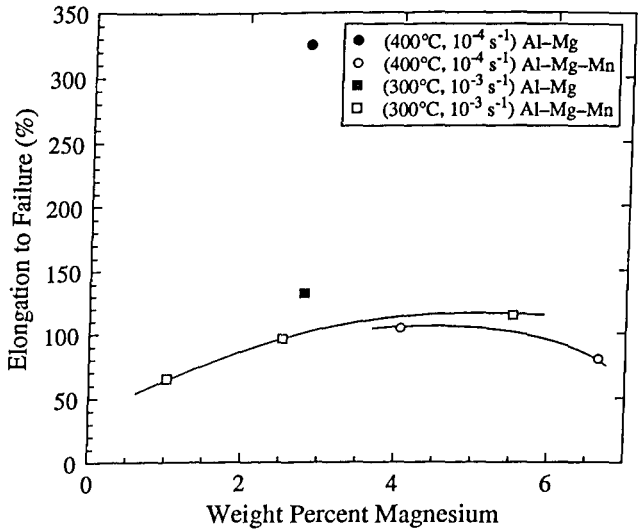

Figure 3: Elongation to failure as a function of Mg concentration for binary Al–Mg and ternary Al–Mg–Mn alloys.

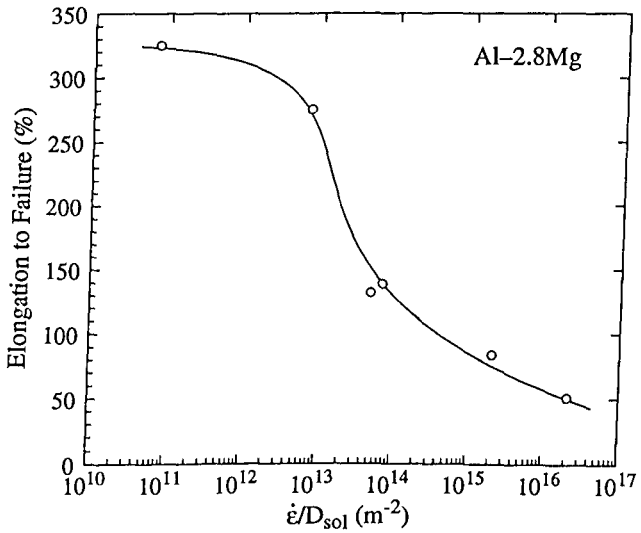

Figure 4: Elongation to failure as a function of diffusion-compensated strain rate for the binary Al–2.8Mg alloy. The diffusion coefficient, D_{sol}, for Mg diffusion in Al is given by $D_0 = 5 \times 10^{-5}$ m^2/s with a corresponding activation energy of $Q = 136$ kJ/mol [8].

References

[1] L. Ivanchev and D. Chavderova, *Proceedings of the International Symposium on Plasticity and Resistance to Metal Deformation*, ed. S. Blecic et al. (Titograd: Fac. Metallurgy, 1986), 348–361.

[2] R. A. Ayres, *Metallurgical Transactions* **10A**, (1979), 849–854.

[3] R. A. Ayres and M. L. Wenner, *Metallurgical Transactions* **10A**, (1979), 41–46.

[4] H. J. McQueen and M. E. Kassner, *Superplasticity in Aerospace*, ed. H. C. Heikkenen and T. R. McNelley (Warrendale: TMS, 1988) 77–96.

[5] O. D. Sherby and P. M. Burke, *Progress in Materials Science* **13**, (1969), 325–390.

[6] D. A. Woodford, *Transactions of the American Society for Metals* **62**, (1969), 291.

[7] M. A. Burke and W. D. Nix, *Acta Metallurgica* **23**, (1975), 793–798.

[8] T. R. McNelley, D. J. Michel, and A. Salama, *Scripta Metallurgica* **23**, (1989), 1657–1662.

[9] E. M. Taleff, D. R. Lesuer, and J. Wadsworth. Manuscript in preparation at Lawrence Livermore National Laboratory.