Mechanical Properties and Cavitation during Superplastic Deformation of Al–Mg Alloys

J. Mukhopadhyay*, A. K. Mallik** and A. M. Rao***

The effect of temperature and strain rate on cavitation during tensile straining in Al–Mg alloys has been studied using precision density measurements, metallography and semi-empirical model of cavity growth. Cavitation occurs throughout the gauge length of the specimens, with most of the cavities being concentrated in regions near the fracture tip. The extent of cavitation has been found to increase with increasing temperature and decreasing strain rate. In fact, the cross-sectional area at fracture increased with increasing level of cavitation and its coalescence. However, cavitation is not found to be optimum around the superplastic temperature, i.e. 673 K and at an initial strain rate of $\varepsilon = 1.4 \times 10^{-3} \text{s}^{-1}$. This is the combination of temperature and strain rate at which the alloys show a maximum elongation to failure and strain rate sensitivity index. Cavity growth at high temperatures may be controlled either by vacancy diffusion or the power law growth process. These two growth mechanisms are examined with reference to Al-Mg alloys. It is found that diffusion growth is favoured at low strains and there is a transition to the power law growth process at a critical cavity radius ($r_c$). The value of ($r_c$) increases with increasing temperature and decreasing strain rate.

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1. Introduction

Cavitation is commonly experienced during high temperature creep exposure. Therefore, several models for cavitation phenomena have been developed to describe creep behavior. Recent investigations have shown that extensive cavitation may also take place during superplastic forming. Hence, it is important to understand the factors which influence cavitation in superplastic alloys in order to minimize their occurrence. Otherwise, the presence of cavities may impose a strong limitation on the effective utilization of the high ductility exhibited during the manufacture of components by superplastic technique.

Many workers have investigated cavitation in several alloys, viz. $\alpha/\beta$ brass, Pb–Sn eutectic, Zn–Al eutectoid, and stainless steel. The present investigation is a detailed study of the cavitation behavior in aluminum-magnesium alloys during superplastic straining. Although the alloys for such an investigation exhibited a reasonable amount of elongations, i.e. 163% and 200%, but they could not be elongated to a fine point fracture stage. Therefore, it was felt that cavitation must be playing a significant role in restricting the overall ductility of these alloys. The purpose of this communication is two fold:

(i) To examine the different parameters responsible for cavitation in these alloys;
(ii) To compare the theoretical prediction with the experimental observations, viz. density measurement and metallographic evidence.

II. Experimental Procedure

1. Materials

Two aluminum alloys had been chosen for this investigation and for both alloys, commer-
cial grade aluminum was used. Alloy A was received through the courtesy of M/S, Indian Aluminum Company, Calcutta, in the form of sheets, whereas alloy B was prepared in the laboratory as follows.

Melting was carried out in a pure graphite crucible in an electrical resistance furnace. Master alloys of Al-Mg and Al-Zr were first prepared and analyzed. These master alloys of Al-Mg and Al-Zr were added to molten aluminum in required proportions. The molten alloy was poured at a temperature of 1003 K into a steel mould, with wall thickness of 15 mm. The diameter of the ingot was 32 mm. The alloys were analyzed and their chemical compositions are given above in Table 1.

The cast ingot of alloy B was homogenized at 703 K for 72 ks and subsequently forged at 723 K to a thickness of 5.40 mm. After forging, it was annealed at 598 K for 900 s, warm rolled (40% reduction) at 473 K and finally cold rolled at room temperature (20% reduction) to a thickness of 2 mm. The specimens for uniaxial test were cut from the 2 mm thick strips. For both alloys A and B, the specimens were solution treated at 623 K for 1.8 ks and 748 K for 2.7 ks, respectively, and subsequently water quenched. The grain sizes were measured by a linear intercept method and reported as spatial grain size, i.e. (1.4 × average linear intercept). The grain sizes of alloys A and B were estimated to be 16 and 26 μm, respectively.

2. Uniaxial tests

Tensile tests for both alloys were carried out on a servo-hydraulic machine (MTS), using specimens of 12 mm gauge length and 4.4 mm width. The specimen thickness was maintained to lie within the limits of 2.0–2.1 mm. A vertical split furnace having a uniform hot zone of 190 mm was used for elevated temperature tests. The temperature was measured by three chromel-alumel thermocouples placed very close to the specimen. Each specimen was held at the test temperature for about 900 s prior to the test. Tests were carried out at different temperatures in the range of 523–723 K and at four different cross head velocities, viz. $3.33 \times 10^{-3}$, $1.67 \times 10^{-2}$, $8.33 \times 10^{-2}$ and $1.67 \times 10^{-1}$ mm s$^{-1}$ which correspond to initial strain rates of $2.8 \times 10^{-4}$, $1.4 \times 10^{-3}$, $6.9 \times 10^{-3}$ and $1.4 \times 10^{-2}$ s$^{-1}$. The parameters such as flow stress, strain rate sensitivity and percentage elongation were evaluated.

3. Cavitation measurement

For both alloys A and B, cavitation was studied by density measurement, metallography and by computation based on two semi-empirical models of cavity growth. For density measurement, the specimens were strained at a constant cross head velocity to a predetermined strain. After straining, the parallel sided gauge lengths and the gauge heads were cut from these specimens and polished. Both gauge length and head portion of the specimens were kept inside the 2 percent collodion solution for 60 s to make the open pores of the surface inaccessible to water. Density measurements were made by hydrostatic weighing in distilled water at a temperature of 299 K. The head portion was used as a reference standard, from where the density decrease in the gauge length was measured and expressed as percentage volume of cavities. The density measurements were carried out on specimens tested at $\dot{\varepsilon} = 1.4 \times 10^{-3}$ s$^{-1}$, to determine the effect of temperature on testing in the range of 523–723 K. Similarly, the effect of strain rate on the volume of cavities was studied at a constant temperature of 673 K over the range of strain rates of $2.8 \times 10^{-4}$–$1.4 \times 10^{-2}$ s$^{-1}$.

III. Experimental Results

1. Mechanical behavior

From a series of true stress-strain curves, an earlier study had shown that the alloys A and B when deformed at 673 K and with a strain rate of $1.4 \times 10^{-3}$ s$^{-1}$ exhibit maximum elongations
of 200 and 163%, respectively. Similarly, the maximum elongation to failure values, for both alloys at the lowest strain rate, i.e. \( \varepsilon = 2.8 \times 10^{-4} \text{s}^{-1} \) and at 673 K were observed around 146 and 138%, respectively. In fact, for all conditions, the alloy A shows more elongation than alloy B. Full details of these are given elsewhere\(^{(7)}\).

2. Cavitation phenomena

Both alloys (A and B) exhibit maximum elongation at 673 K and with a strain rate of \( 1.4 \times 10^{-3} \text{s}^{-1} \). Hence, it seemed to be appropriate to study the cavitation phenomena quantitatively at a combination of these two parameters. Therefore, density measurements based on these parameters were made on both of these alloys to determine the overall level of cavitation.

It is observed that cavitation increases with increase of temperature and decrease of strain rate. The cavitation is more pronounced in alloy B than alloy A over the entire range investigated.

3. Comparison of theoretical prediction with experimental data

Based on pioneering work by Hull and Rimmer\(^{(9)}\), subsequently developed by several investigators\(^{(10)-(14)}\), the cavity growth rate given by Miller and Langdon\(^{(15)-(16)}\) was taken into consideration for the present work.

For diffusion growth, the rate of change of cavity radius \( r \) with \( \varepsilon \) is given by\(^{(10)}\),

\[
\frac{dr}{d\varepsilon} = \frac{1}{\varepsilon} \left[ \frac{\Omega \delta D_{gb}(\sigma - 2\gamma / r)}{2KTr^2[\ln (a/2r) - 1/2]} \right] \tag{1}
\]

where \( \ln (a/2r) - 1/2 = 1 \) \( \tag{2} \)

For power law growth\(^{(13)}\),

\[
\frac{dr}{d\varepsilon}_p = r - \left[ \frac{3\gamma}{2\sigma} \right] \tag{3}
\]

\( \varepsilon \) = strain rate, \( \Omega \) is the atomic volume, \( \delta \) is the width of grain boundary, \( D_{gb} \) is the grain boundary diffusion coefficient, \( \sigma \) is the applied stress, \( \gamma \) is the surface energy, \( K \) is the Boltzmann’s constant, \( T \) is the absolute temperature and \( a \) is the cavity spacing.

The appearance of cavities is different for these two growth processes. It is reported that diffusion growth leads to essentially spherical cavities, preferentially located on grain boundaries lying approximately perpendicular to the stress axis, whereas power law growth leads to cavities which are elongated in the direction of tensile stress. In general, power law growth is favored at high strain rates and diffusion growth dominates at low strain rates.

For determining the growth rate for alloy A, a computer program has been developed in the present investigation on the basis of these two equations, i.e. eq. (1) and eq. (3). Calculations were performed by putting \( \Omega = 1.66 \times 10^{-29} \text{ m}^{3} \), \( \delta = 2b = 5.72 \times 10^{-10} \text{ m} \), \( D_{gb} = 0.1 \times 10^{-4} \exp (0.-60290 \times 10^{5} RT) \text{ m}^{2} / \text{s} \), \( \gamma = 1.1 \text{ J/m}^{2} \) and \( R = 8.31 \text{ Jmol}^{-1} \text{K}^{-1} \). Figures 1(a) and (b) show the plots of \( dr/d\varepsilon \) versus \( r \) for both diffusion and power law growth processes. The corresponding flow stress values at 673 K were estimated to be 44 and 138.28 MPa\(^{(7)}\).

It can be seen from Fig. 1(a) that diffusion growth at \( \varepsilon = 2.8 \times 10^{-4} \text{s}^{-1} \) becomes a slower process, when the radius of cavity reaches a critical value of \( r_c = 2.5 \mu \text{m} \). At larger radii, i.e. beyond 2.5 \( \mu \text{m} \), growth takes place mainly by power law mechanism. Figure 1(b) suggests that power law growth rate at \( \varepsilon = 1.4 \times 10^{-2} \text{s}^{-1} \) becomes a dominant process for cavity radius greater than 1 \( \mu \text{m} \). Therefore, the critical cavity radii were computed to lie in the range of 1.0–2.5 \( \mu \text{m} \) for the conditions of strain rate and temperature used in Figs. 1(a) and (b).

For determining the critical cavity radii theoretically for alloy A at different temperatures in the range of 523–723 K and strain rates \( 2.8 \times 10^{-4} \text{s}^{-1} - 1.4 \times 10^{-2} \text{s}^{-1} \), the flow stress values were taken from the steady state region of true stress and strain curves\(^{(7)}\). The four sets of curves in Fig. 2 and the five sets in Fig. 3 show the linear relationship, where critical cavity radius increases with increase of temperature and decrease of strain rate.

4. Variation of mechanical properties with cavitation

To correlate the mechanical properties with the level of cavitation, data on elongation to failure and strain rate sensitivity index (m)
have been superimposed, along with volume of cavities versus temperature at \( \dot{\varepsilon} = 1.4 \times 10^{-2} \text{ s}^{-1} \) in Figs. 4(a) and (b). Around 673 K, it is observed that both elongation to failure and strain rate sensitivity index (m) are maximum, i.e. 200%, 0.34 for alloy A and 163%, 0.31 for alloy A at different temperatures.
alloy B. However, the volume of cavities continues to increase with increase in temperature, i.e. up to 723 K (the highest temperature used for this investigation).

For alloy A, the elongation to failure and the critical cavity radius, i.e. $r_c$ ($r_c$ was determined by semi-empirical model of cavity growth which is discussed in section 3) were superimposed on volume of cavities against log initial strain rate in Fig. 5. It can be seen from this figure that cavitation is not maximum at intermediate strain rate $\varepsilon = 1.4 \times 10^{-3} \text{s}^{-1}$, where alloy A exhibits extensive deformation, i.e.
200%. Similarly, for alloy B in Fig. 6, the maximum elongation around 163% is found to occur at intermediate strain rate. Cavitation increases continuously with decrease of strain rate and it becomes more pronounced at $\dot{\varepsilon} = 2.8 \times 10^{-4} \text{ s}^{-1}$, the strain rate up to which the investigation was carried out.

5. Metallography

Same specimens which were used for density measurements were later used for metallographic examination. Metallographic evidence shows that cavitation occurs throughout the gauge length of the fractured samples, with most of the cavities being located in the vicini-
ty of the fractured surfaces.

At the lowest strain rate, i.e. $\varepsilon = 2.8 \times 10^{-4}$ s$^{-1}$, the extent of cavitation for both alloys increases significantly. This is demonstrated in Figs. 7(a) and (b). Extensive cavitation is visible close to the point of fracture, since cavities being concentrated in areas of high strain\(^2\). Apart from that, there is also a tendency for cavity interlinkage. However, overall level of cavitation appears to be less extensive for alloy A as compared to alloy B.

Figures 9(a) and (b) show the cavitation in both alloys tested at strain rate of $1.4 \times 10^{-2}$ s$^{-1}$. As compared to previous strain rate, i.e. $\varepsilon = 2.8 \times 10^{-4}$ s$^{-2}$, the cavities are less extensive here. Metallographic studies thus show that the extent of cavitation decreases with increase of strain rate.

A similar sequence was also observed at different temperatures in the range of 523–723 K. Cavitation is more at the fracture tip at all temperatures and it decreases continuously within the gauge length at a point remote from the fracture tip. The extent of cavitation increases with increase of temperature. Therefore, a good correlation exists between the volume of cavities determined experimentally by density measurements and with those observed under the microscope.

IV. Discussion

Ayres\(^{19}\) has indicated two possible mechanisms which are responsible for deformation in aluminum-magnesium alloys, viz.

i) dynamic strain ageing

ii) dynamic recovery

The dynamic strain ageing is found to be operative at ambient temperature, whereas dynamic recovery is the sole softening mechanism at elevated temperature.

At the lowest strain rate, i.e. $\varepsilon = 2.8 \times 10^{-4}$ s$^{-1}$, as the temperature is increased, the amount of softening goes up due to dynamic recovery. Therefore, the elongation to failure is expected to be higher, but the opposite was found in this case. This probably indicates the effect of cavitation on elongation to failure.

1. Effect of temperature on cavitation

In the temperature range of 500 to 600 K, the elongation to failure values for both alloys in Figs. 4(a) and (b) are very low. This is possible, since the strain rate sensitivity values which resist necking at this stage are also very low, nearly 0.1. Therefore, a neck develops.
during the initial straining of the specimen. With further straining, deformation is concentrated in the neck region which ultimately leads to premature failure.

In the temperature range of 600–700 K, the cavitation also increases as compared to previous range; therefore, the ductility should come down instead of going up. This is not so, since the alloys possess the highest strain rate sensitivity values i.e. 0.34 for alloy A and 0.31 for alloy B. It is believed that these high values of strain rate sensitivity will resist macroscopic necking, which occurs due to cavitation and its coalescence.

At high temperature i.e. 700 K and above, the maximum elongation to failure values for both alloys steadily drops, despite reasonably high strain rate sensitivity values achieved, i.e. 0.24 for alloy A and 0.28 for alloy B. It appears that cavity growth and interlinkage of cavities is of major significance in this range for limiting the ductility. Furthermore, the strain rate sensitivity value for alloy B is more than alloy A, therefore, elongation to failure in case of alloy B is expected to be higher than alloy A. In this case, the reverse is true, because cavitation observed for alloy B is more than that of alloy A. This is confirmed by comparing the Figs. 7(a) and (b). Hence, cavitation ultimately results in reducing the ductility of alloy B.

As the alloys contain magnesium, vacancy concentration in these alloys increases due to magnesium solute\(^{20}\). If the temperature is raised, the existing cavity nuclei start growing and this growth is enhanced by vacancy diffusion which ultimately leads to an increase in cavitation with temperature. Thus, cavitation increases with increase of temperature in Figs. 4(a) and (b).

In some materials, an increase in temperature also leads to a decrease in cavitation\(^{21}\). However, this has not been observed in the present investigation.

2. **Effect of strain rate on cavitation**

The extent of cavitation decreases with increase of strain rate. Because, the higher stresses associated with increased strain rate could nucleate increased number of cavities, which are able to grow by vacancy mechanism for a time which is inversely proportional to the strain rate. Therefore, the cavity cannot grow with reduced time. Besides that, void growth is less probably due to the decreased contribution of grain boundary sliding at high strain rates. In fact, the growth is more pronounced at the lowest strain rate, i.e. \( \dot{\varepsilon} = 2.8 \times 10^{-4} \text{s}^{-1} \), where more time is available for a cavity to grow. In the process, cavities observed are often large and irregular in shape, shown in Fig. 7. This is consistent with increasing importance of the diffusional growth of cavities at very low strain rates. Apart from that, there is also a tendency for cavity interlinkage for both alloys in Figs. 7(a) and (b). It appears that the fracture occurred at low strain rate, by a combination of external necking and interlinkage of voids by an intergranular void sheet process\(^{22}\).

Figures 7 to 9 also show a typical example of cavities observed within the gauge length at a temperature of 673 K and two different strain rates of \( 2.8 \times 10^{-4} \) and \( 1.4 \times 10^{-2} \text{s}^{-1} \), respectively. The tensile axis is vertical for all photomicrographs. Here, the smallest cavity in the range of (2–5 \( \mu \text{m} \)) away from the fracture tip has been taken into account, since they are more representative. If the coarser cavities near to the fracture tip are taken into consideration, then the cavities observed experimentally will be significantly larger than the critical cavity size (2.5 \( \mu \text{m} \)) obtained through computation in Fig. 1(a). It is thus apparent that interlinkage of cavities plays an important role for additional cavity growth. Subsequently, if void growth is strain controlled, then the void volume fraction would be found to increase exponentially with strain. This has been observed in other alloys such as \( \alpha/\gamma \) micro duplex stainless steel and several aluminum base alloys\(^{21}\).

Figure 1(b) suggests that power law mechanism becomes the dominant process for cavity radii greater than 1 \( \mu \text{m} \), while deformed at \( \dot{\varepsilon} = 1.4 \times 10^{-2} \text{s}^{-1} \). This is consistent in Figs. 9(a) and (b), which show the importance of power law growth. Here, the cavities are small in size and their sizes, in fact, lie in the range of (0.4–1.0 \( \mu \text{m} \)). Whereas at a lower strain rate,
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i.e. \( \varepsilon = 2.8 \times 10^{-4} \text{ s}^{-1} \), a corresponding increase in the value of critical cavity radius \( r_c \) is observed. It is thus apparent that diffusion growth increases in importance (therefore, the general trend is consistent with the experimental observation of a transition from predominantly power law growth at fast strain rates, to predominately diffusion growth at low strain rates).

V. Conclusions

(1) Cavitation is observed throughout the gauge length of fractured samples for both alloys A and B and it is found to increase monotonically with temperature. This happens probably due to enhanced cavity growth by vacancy diffusion at high temperature.

(2) Cavitation decreases with increase of strain rate. It thus appears that at high strain rate, reduced time is available for a cavity to grow and subsequently interlink to make a larger pore size.

(3) Computation shows that the critical cavity radius for change-over of growth mechanism from a diffusional growth to a power law growth for alloy A increases with decreasing strain rate. At 673 K, the critical cavity radii are in the range of 1 \( \mu \text{m} \) and 2.5 \( \mu \text{m} \) for strain rates, \( \varepsilon = 1.4 \times 10^{-2} \) and \( 2.4 \times 10^{-4} \text{ s}^{-1} \), respectively.

(4) There are significant differences in the cavitation behaviour between the two temperature ranges, i.e. 500–600 K and, 700 K and above. In the former range, the cavitation is less, whereas in the latter range, macroscopic necking occurs and it appears that growth and interlinkage of cavities is of major importance in this range, for limiting the ductility.

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REFERENCES

(1) D. M. R. Taplin and R. F. Smith: Fracture 77, Vol. 2, ed. by D. M. R. Taplin, University of Waterloo, Ontario, p. 541.
(2) J. W. D. Patterson and N. Ridley: J. Mater. Sci., 16 (1981), 457.
(3) C. W. Humphries and N. Ridley: J. Mater. Sci., 12 (1977), 851.
(4) N. Ridley and D. W. Livesey: Fracture 77, Vol. 2, ed. by D. M. R. Taplin, University of Waterloo, Ontario, p. 533.
(5) M. I. Mohamed, F. A. Mohamed and T. G. Langdon: J. Mater. Sci., 14 (1979), 2913.
(6) C. I. Smith, B. Norgate and N. Ridley. J. Mat. Sci., 10 (1976), 182.
(7) J. Mukhopadhyay: Ph. D Thesis, Dept. of Met. Engg., Indian Institute of Technology, Bombay, 1982.
(8) J. Mukhopadhyay, A. M. Rao and A. K. Malik: Trans. IIM., 38 (1985), 389.
(9) D. Hull and D. E. Rimmer: Phil. Mag., 4 (1959), 673.
(10) M. V. Speight and J. E. Harris: Met. Sci. J., 1 (1967), 83.
(11) M. V. Speight and W. Beere: Met. Sci., 9 (1975), 190.
(12) R. Raj and M. F. Ashby: Acta Met., 23 (1975), 653.
(13) J. W. Hancock: Met. Sci., 10 (1976), 319.
(14) F. A. McClintock, J. Appl. Mech. 35 (1968), 363.
(15) D. A. Miller and T. G. Langdon: Met. Trans., 10A (1979), 1869.
(16) D. A. Miller and T. G. Langdon: Trans. JIM, 21 (1980), 123.
(17) M. F. Ashby: Acta Met., 20 (1972), 887.
(18) H. Jones: Metal Sci. J., 5 (1971), 15.
(19) R. A. Ayres: Met. Trans., 8A (1977), 487.
(20) E. C. W. Perryman: Trans. AIME, 8 (1956), 1247.
(21) J. Pilling and N. Ridley: Res. Mechanica., 23 (1988), 1-33.
(22) H. Ishikawa, D. K. Bhat, F. A. Mohamed and T. G Langdon: Met. Trans., 8A (1977), 523.