

Static creep behaviour of Al–Zn–Mg and Al–Zn–Mg–Cu alloys

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The creep behaviour of Al–Zn–Mg (7039) and Al–Zn–Mg–Cu (7075) alloys is evaluated at elevated temperatures ($443 \leq T \leq 533$ K and $483 \leq T \leq 563$ K) under constant stresses between 49 and 123 MPa, respectively, in a custom-built creep testing facility. The measured activation energies of these alloys are 172–185 kJ mol⁻¹ and 248–272 kJ mol⁻¹. As the stress increases, the activation energy in both cases decreases due to the high density of dislocations. The average exponent values of these alloys are 7 and 9. The microstructure observation reveals that the dominant fracture mode of 7039 alloy is intergranular and that of 7075 alloy is transgranular.

1. Introduction

7000 series aluminium alloys are emerging as the candidate material of choice for aerospace and structural applications because of their excellent physical and chemical properties, such as high strength, low density, stress corrosion cracking resistance, high-temperature capability, machinability, etc.

Ludtka and Laughlin [1] examined the fracture mode and toughness of the 7075 aluminium alloys in the yield strength range of 496 to 614 MPa and found that toughness decreased as the yield strength increased with increasing solute content, (Zn + Mg). However, the 7075 Al alloy in the present work has 5.98 wt % Zn and 2.62 wt % Mg, i.e. the so called low solute alloy. From transmission electron microscopy (TEM) and X-ray observations, they reported that the fracture mode of the low solute alloy was a transgranular dimpled rupture due to the fine Cr-dispersoids (E-phase). However, the fracture mode changed to intergranular due to void coalescence at the grain boundary dispersoids with increasing solute content. Sherby *et al.* [2] investigated the activation energy of high-purity aluminium (99.996%) in the temperature range 77 to 880 K under constant stress conditions. They found that the activation energy at temperatures between 500 and 880 K was 149 kJ mol⁻¹, and that the activation energy for self-diffusion of aluminium, and between 250 and 375 K was 116 kJ mol⁻¹. The activation energy for creep of pure aluminium above 0.5 T_M was not a function of the temperature. Drury and Humphreys [3] observed, from a compression test of Al–5% Mg alloy, that the mechanical behaviour at 80–90 MPa showed a transition from viscous glide (pure metal type) to microband formation (alloy type). Rosler *et al.* [4] examined the mechanical prop-

erties of dispersion-strengthened aluminium alloys and found that dispersoid–dislocation interaction was key point to excellent creep properties. Rios [5] reported that there was a limiting grain size above 798 K due to the insoluble dispersoids. Only the insoluble dispersoids are acting on the drag force because higher temperatures increase the solubility. However, the creep behaviour of the 7000 series alloys, which exhibit the most important properties for aerospace application, has not been studied in detail under operating environments.

The main objective of the present paper is to report on the static creep behaviour of a 7000 series alloy, i.e. steady state creep rate, stress exponent and activation energy for creep, and Larson–Miller parameters, as well as determine the creep deformation mechanism.

2. Experimental procedure

The 7039 Al alloy was prepared from high purity metals, melted in a high frequency induction furnace under an argon atmosphere and cast. The plate-type ingot was then cold rolled down to a thickness of 1 mm. The sample preparation procedure is similar to that reported by Ryum [6]. Creep specimens with a gauge length of 25 mm were prepared along the rolling direction by the electric-discharge method (EDM) as shown in Fig. 1. The as-received 7075 Al alloy is a plate, supplied by the Korean Air Force. The as-received plate was initially cut by laser and then prepared by the EDM method. All specimens were then mechanically polished with 600, 800, and 1200 grit SiC papers before the experiment.

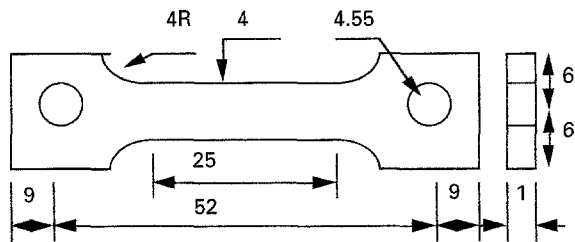


Figure 1 Dimensions of the creep specimen (unit: mm).

TABLE 1 Chemical composition of 7039 and 7075 Al alloys (wt %)

Alloys	Al	Zn	Mg	Cu	Cr	Mn
7039	Bal.	4.3	1.7	0.006	0.001	0.001
7075	Bal.	5.98	2.62	1.87	0.22	0.01

Both alloys were prepared by the two-stage ageing process. The 7039 specimens were homogenized at 723 K for 4 h, water quenched to room temperature, and then naturally aged for 7 days. The 7075 specimens were homogenized at 760 K for 4 h and quenched in water [6]. Then the T6 temper was carried out prior to testing to improve the strength and stress-corrosion cracking resistance [7]. The average linear intercept grain sizes, determined by an optical microscope, were 250 μm (7039) and 100 μm (7075). The chemical compositions (in wt %) of the major alloying elements in these materials are given in Table I.

All creep tests were performed in a lever-beam type test facility as developed by Andrade and Chalmers [7]. The cam lever was adjustable so as to account for different initial specimen lengths by changing the point of the initial load [8]. Therefore, only one Duralumin cam was employed in this work. The specimens were held in Inconel 713C grips with stainless steel bolts and nuts inserted into the chamber and heated by a nichrome heating unit (chamber) enclosed with a quartz tube to obtain a uniform temperature gradient. The specimen temperature was then sensed by a thermocouple attached to the centre of the gauge section and measured by a digital temperature meter to a precision of ± 3 K. The displacement of the specimen was detected and recorded by using a linear variable differential transducer (LVDT) attached to the grip with an accuracy of ± 0.0005 mm. The configuration of the present test machine and details of the experimental procedure are described elsewhere [3].

Above 340 K, the strength of Al alloys must be taken into account. The airplane body experiences a 80% strength reduction down to 24 from 147 MPa when the temperature increases up to 488 K at a speed of Mach 2.5. Therefore, the 7039 Al alloy creep tests were performed in air under conditions of constant stress in temperatures between 443 and 543 K, with

stresses between 49 and 123 MPa. The 7075 Al alloy was also tested in air under conditions of constant stress in temperatures between 483 and 563 K and stresses between 49 and 123 MPa. The fracture surface of the tested specimens was then examined with an optical microscope and a scanning electron microscope (SEM, Jeol JSM-840A) to determine the fracture mode.

3. Results and discussion

The effect of stress on the creep rate of 7039 and 7075 alloys is shown in Figs 2 and 3, log-log plotted by using the least squares method. The data show straight lines at all temperatures. The results indicate that, as the temperature increases up to 533 K, the stress exponent in both cases decreases as expected with decreasing stress. The average stress exponent values (7 and 9) are within the power law breakdown

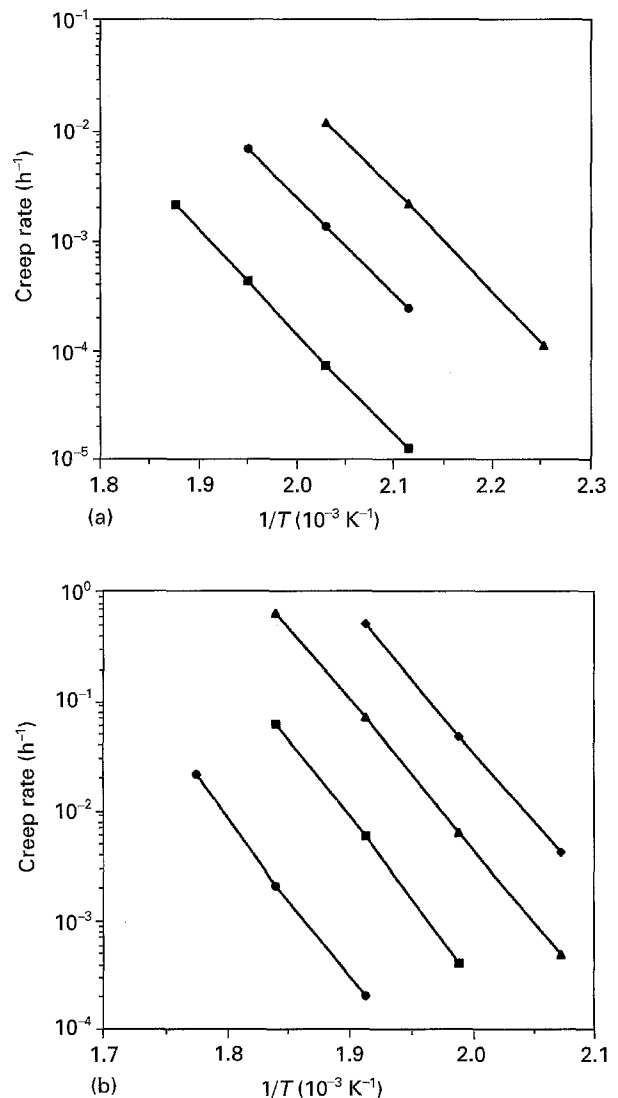


Figure 2 Effect of temperature on the steady state creep rate for (a) 7039 and (b) 7075 Al alloy in static creep. Key: (a) ■ 185 kJ mol^{-1} , 49 MPa; ● 179 kJ mol^{-1} , 74 MPa; ▲ 172 kJ mol^{-1} , 98 MPa; (b) ■ 269 kJ mol^{-1} , 55 MPa; ● 277 kJ mol^{-1} , 49 MPa; ▲ 264 kJ mol^{-1} , 98 MPa; ◆ 249 kJ mol^{-1} , 123 MPa.

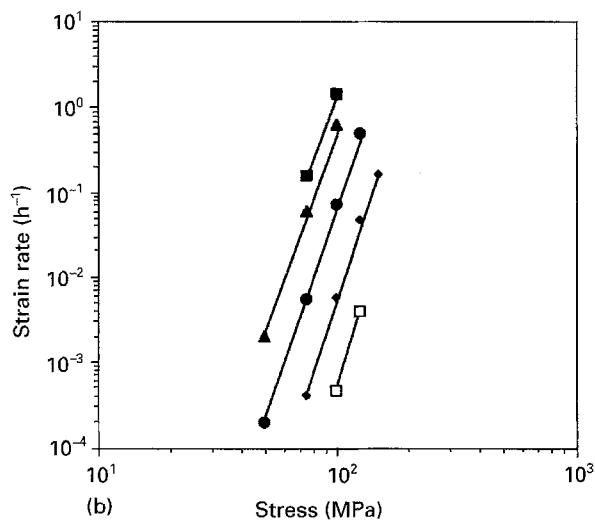
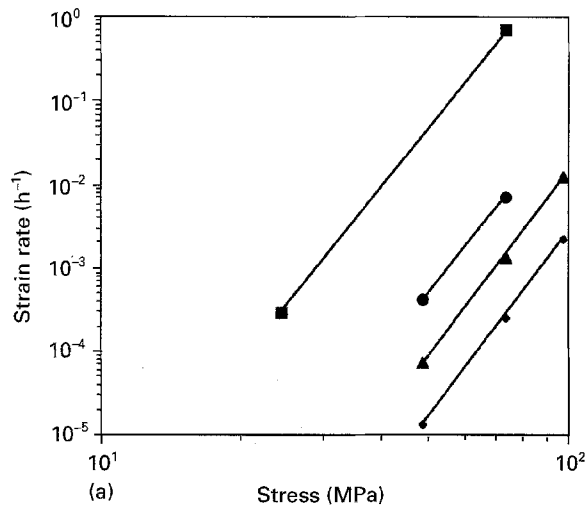


Figure 3 Effect of applied stress on the creep rate for (a) 7039 and (b) 7075 Al alloy in static creep. Key: (a) \blacklozenge 473 K, $n = 7.3$; \blacktriangle 493 K, $n = 7.1$; \bullet 513 K, $n = 6.88$; \blacksquare 533 K, $n = 6.43$; (b) \blacksquare 563 K, $n = 8.4$; \blacktriangle 543 K, $n = 8.5$; \bullet 523 K, $n = 8.8$; \blacklozenge 503 K, $n = 9.2$; \square 483 K, $n = 9.6$.

regime [3] which represents the behaviour of alloys deforming at high stresses. However, the dislocation density decreases with increasing temperature. In fact, it corresponds to a reduction in the applied stress. The high density of dislocations during ageing and plastic deformation also makes the movement of a dislocation line in a matrix easier by climb, which results in dislocation multiplication. The average stress exponent values of 7 and 9 in these alloys implied that the creep was controlled by recovery rather than viscous glide due to the breakage of dislocations from their solute atmosphere (Zn, Mg) as reported by Drury and Humphreys [3].

The activation energy for creep was determined from an Arrhenius-type equation. The activation energy was measured from the slope of steady state creep as depicted in Figs 4 and 5. The average activation energy for static creep of the 7039 Al alloy was about 180 kJ mol^{-1} . In the 7075 Al alloy, it was about 263 kJ mol^{-1} . As the stress increases, the activation

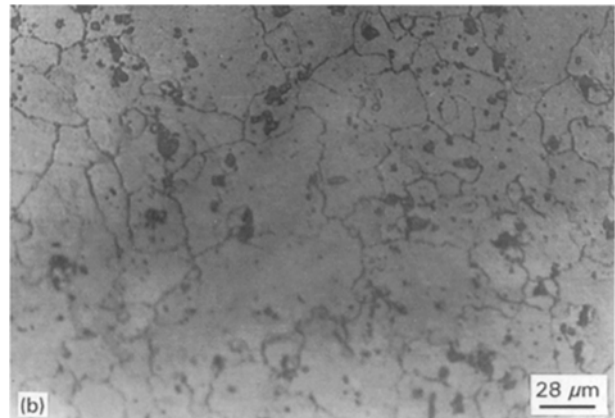
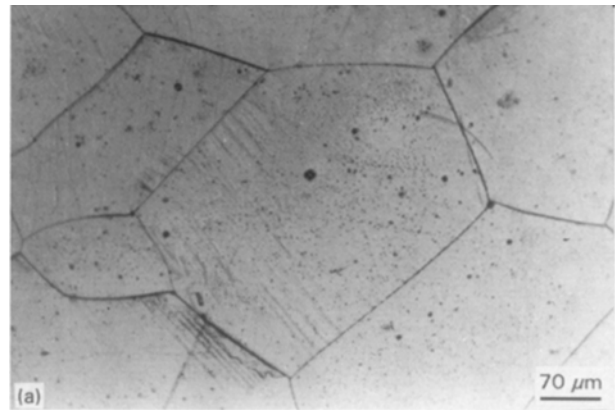


Figure 4 Optical micrographs of (a) 7039 and (b) 7075 Al alloys showing different fracture modes ($\times 140$).

energy in both cases decreases as expected due to the high density of dislocations. It is believed that the high density of dislocations with increasing stress contributes to the low activation energy value due to the ease of diffusion. Adabbo *et al.* [9] found that the activation energy of a 7475 Al alloy at 673 to 788 K was about 141 kJ mol^{-1} , which was the same as that for lattice diffusion of Al. The measured value was higher than that of lattice diffusion. These measured activation energy values may be the sum of those for lattice diffusion and for solution of precipitates (MgZn_2 , Al_2CuMg , etc.) [10].

The Larson–Miller parameter in static creep was employed to predict the long-term rupture behaviour. The constant value was determined by the graphical method using the Larson–Miller parameter. The constant value for the 7039 Al alloy was 8.6 and 14 for the 7075 Al alloy. The lower value of the constant indicated a longer rupture time. This phenomenon may be explained by the fact that creep rates decrease with increasing stress and temperature due to the effect of dislocation recovery.

The creep rate is a strong function of temperature, stress, and structure factor. The phenomenological equations for steady state creep in 7039 and 7075 Al alloys were obtained over the temperature region employed in the present study with a computer simulation. The equations of steady state creep were derived

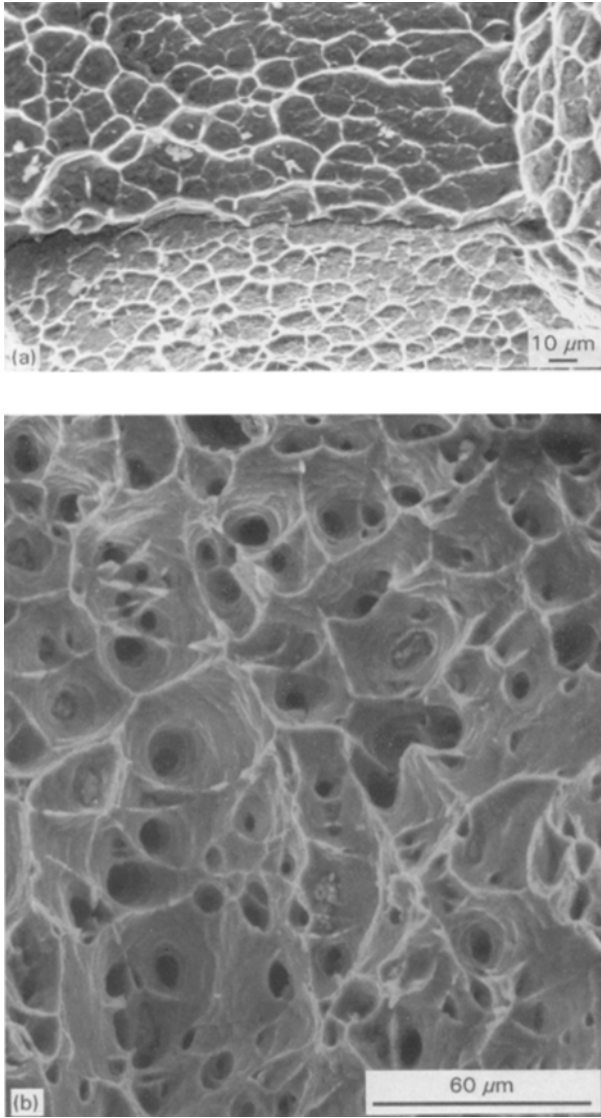


Figure 5 SEM micrographs of (a) 7039 at 98 MPa and 493 K and (b) 7075 Al alloy at 74 MPa and 563 K.

from an Arrhenius equation and creep parameters (strain, temperature, stress, structure factor). The creep rate equation for the 7039 Al alloy is,

$$\dot{\epsilon} = \exp[-5.519 \times 10^{-4} \sigma + 2.33 \times 10^{-2} T - 0.68 \sigma + 18.295] \times \sigma^{-0.0142T + 10.18} \exp\left[\frac{(-6\sigma + 47.8)1000}{RT}\right] \quad (1)$$

where $\dot{\epsilon}$ is the static creep rate, σ is the stress (MPa), T is absolute temperature (K), and R is the gas constant. The creep rate equation for the 7075 Al alloy is described as follows:

$$\dot{\epsilon} = \exp[(4.068 \times 10^{-3} \sigma + 9.8 \times 10^{-4} T - 2.934 \sigma + 44.15] \times \sigma^{-1.675 \times 10^{-2} T + 17.6} \exp\left[-\frac{(-0.864 \sigma + 70.56)1000}{RT}\right] \quad (2)$$

It was found that the experimental observations in the present study were consistent with the above equations.

The fracture surfaces of the tested specimens were examined with an optical microscope and a SEM. The optical micrograph of the 7039 crept specimens (true strain, $\epsilon_t = 0.03$) as shown in Fig. 4 indicated that traces of slips existed across the grain boundary and therefore, grain boundary accommodation of the slip was inevitable. The strain accommodation produces a precipitate-free zone (PFZ) along the adjacent grains because the T6 temper aluminium alloys always provide precipitates [11]. Then, the grain boundary and slip interactions may initiate the formation of voids due to strain localization and final intergranular failure. Fig. 5 shows that typical round-type cracking occurred along the grain boundary due to the diffusion of vacancies and grain boundary sliding. Observation of the SEM micrograph, however, reveals that the grain boundary fracture with an irregular shape is the result of shear stress. The shear stress causes grain boundary sliding to occur. The white lines near the grain boundary oriented 45 degrees to the tensile axis, due to the maximum shear stress, represent slip systems with several specific orientations as reported by Watanabe [12]. He mentioned that grain boundary irregularities were the primary sites for stress concentration and crack nucleation and the number of cavities depended only on the amount of sliding in creep. However, no zigzagged cracking as an indication of grain boundary migration was observed. The fracture mode of the 7075 alloy was completely different from that of the 7039 alloy because the 7075 alloy has a much higher Cr content as shown in Table I. It ($\epsilon_t = 0.02$) failed predominantly by a transgranular mode with a very fine population of dimples of uniform size inside the grain, as shown in Fig. 5, due to the void formation at the Cr-dispersoids (E-phase), as reported by Ludtka and Laughlin [1]. These dispersoids existing within the grain are acting as a void nucleation source which finally leads to ductile fracture within the grain. It is observed that the Cr-dispersoids across all the grains are distributed uniformly, as shown in Fig. 5.

4. Conclusions

1. The stress exponent value in the 7039 Al alloy was 7, and 9 in the 7075 Al alloy. The exponent values were above the power law regime. They decreased with increasing temperature.
2. The measured activation energy values were 179 kJ mol⁻¹ for the 7039 Al alloy and 260 kJ mol⁻¹ for the 7075 Al alloy. These values were both higher than those for lattice diffusion.
3. The constant value of the Larson–Miller parameter in the 7039 Al alloy was 8.6 and 14 in the 7075 Al alloy.
4. The fracture mode in static creep of 7039 alloys was intergranular. The grain boundary–slip interaction due to shear stress and PFZ were responsible for the formation of cavities at the grain boundaries. However, the 7075 alloys failed inside the grain as expected, i.e. the so-called transgranular fracture mode. It is believed that each void is initiated, grown, and fails at the Cr-dispersoids, respectively.

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*Received 27 May 1994
and accepted 9 November 1995*