Creep Deformation Mechanisms in Coarse-Grained Solid Solution Mg Alloys

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Creep deformation behavior of coarse-grained Mg-Al based solid solution alloy (AZ31) was studied in a wide strain rate range of 2 × 10^{-5}–7 × 10^{-2} s^{-1} at temperature range of 573–673 K. Viscous glide controlled creep (VGC), dislocation climb creep (DCC) and power law breakdown (PLB) showed up in order with increasing stress as the flow rate-controlling process. From the former results for Mg-Al and Mg-Al-Zn alloys, the creep mechanisms of VGC and DCC in Mg and Mg alloys are confirmed. Moreover, several theories presented for DCC and VGC are applied to convey the creep mechanisms in Mg and Mg alloys by analytical way. Transitions in Mg alloys are analyzed also by comparing experimental results and theories. Alloying effects on creep strength and transitions of deformation mechanism are analyzed.

(Received October 27, 2003; Accepted February 16, 2004)

Keywords: creep, magnesium alloys, high temperature deformation, transition

1. Introduction

Since half a century, there have been progressive researches on creep behavior of solid solution materials. Those have been related to the deformation mechanisms that can be classified into two types; class II (also called as class M, which follows dislocation climb model (hereafter, DCC)) and class I (also called as class A, which involves viscous glide motion of dislocations dragging solute atmospheres).1–3) Weertman4) developed a theoretical model for power-law creep by assuming that dislocation climb velocity controls the deformation rate. Another creep mechanism for DCC was proposed by Sherby5) in a phenomenological way based on numerous experimental data obtained from a variety of metallic materials. For viscous glide controlled creep (hereafter, VGC), three theoretical models are available, which had been developed by Weertman,6) Friedel,7) Takeuchi and Argon,8) respectively. Friedel7) proposed a theory that dislocations break away from their solute atmospheres when applied stress is high enough. This concept of breakaway stress has been successfully applied to explain the transition from VGC to DCC at high stresses in many Al-Mg and In-based solid solution alloys.9)

Deformation mechanisms of Mg alloys recently attract attention since magnesium is being pursued as the next generation metal used for structural application due to its ultra-lightness. Vagarali and Langdon10,11) had presented comprehensive studies on elevated temperature creep mechanisms of pure magnesium and Mg-0.8%Al over the wide range of strain rate (10^{-8}–5 × 10^{-3} s^{-1}) and temperature (473–820 K). Three regions showing different deformation behavior could be distinguished in both pure Mg and Mg-0.8%Al alloy, depending on temperature and stress ranges. Non-basal slip was dominant at high temperatures (above 600–750 K), while DCC or VGC was dominant at relatively low temperatures (from 523 to 623 K). Unlike in the pure Mg where DCC is dominant over the entire stress range in this temperature range, VGC emerges in the Mg-0.8%Al alloy as the rate-controlling process at low stresses. This result attests that Mg-Al alloy is a Class I type solid solution alloy. Then, there followed several researches about creep transitions on coarse grained Mg and Mg alloys.12–15)

Kim et al.15) studied deformation mechanisms of fine-grained AZ31 and AZ61 Mg alloys at elevated temperatures. They constructed deformation mechanism maps using the constitutive equations derived from the experimental results. They could observe grain boundary sliding (GBS), DCC and power law breakdown are dominant deformation mechanisms in the fine-grained Mg alloys. VGC was not observed in their materials but according to the map they constructed it would appear in low stress range at large grain size. Watanabe et al.16) indeed reported the appearance of VGC in coarse grained AZ31 Mg alloy (grain size, d~ 130 μm, where d = 1.74 × L, L is linear intercept grain size) in low stress range at temperatures 598–673 K.

In this report, deformation behavior of a commercial AZ31 alloy of large grain size was investigated and compared with those of pure Mg and other Mg-Al, Mg-Al-Zn alloys. Effect of alloy concentration on creep behavior of Mg is especially focused on.

2. Analysis

In the former work,15) commercial AZ31 alloy sheet (Mg-3 mass%Al-1 mass%Zn, d ~ 150 μm) by hot extrusion was prepared for the experimental data. The tensile specimens were extracted from the as-received sheet by wire electro-discharging machining with geometry of gauge length of 5 mm, width of 5 mm and thickness of 2.5 mm. In order to obtain the strain rate-stress relationship, strain rate change test (SRC test) were carried out in temperature range of 573–648 K and initial strain rates from 1 × 10^{-4} to 1 × 10^{-2} s^{-1}. For the Watanabe et al.’s data,16) similar commercial AZ31 alloy was used for tensile elongation-to-failure tests in temperature range of 573 to 673 K and constant strain rates from 1 × 10^{-5} to 1 × 10^{-3} s^{-1}. The true strain rate vs. true stress data were determined at a fixed strain of 0.1.

Figure 1 shows true stress-true strain rate data for the
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Fig. 1 Strain rate vs. stress curves of coarse-grained AZ31 including Watanabe et al.’s result.\(^{15,16}\)

current AZ31 Mg alloy plotted in log strain rate vs. log \(\sigma/E\) at temperatures in range between 573 K and 673 K. The data of Watanabe et al.\(^{16}\) are also plotted. Since Watanabe et al.’s datum for \(10^{-3}\) s\(^{-1}\) falls right onto the curve for the current data and the datum for \(10^{-4}\) s\(^{-1}\) resides on the extension of the current curve to lower stress range, it can be concluded that the two materials are virtually identical. Three regions are distinguishable to be associated with different deformation mechanisms when viewed in the entire strain-rate range ranging from \(10^{-5}\) to \(10^{-1}\) s\(^{-1}\); one exhibits \(n = 3\) in low stress regime (\(\sigma/E = 1.6 \times 10^{-4} - 7.8 \times 10^{-4}\)), another does \(n = 5\) in intermediate stress regime (\(\sigma/E = 7.7 \times 10^{-4} - 1.4 \times 10^{-3}\)) and the other does \(n \geq 7\) in high stress regime (\(\sigma/E = 1.4 \times 10^{-3} - 2.4 \times 10^{-3}\)).

There has been developed a constitutive equation interpreting high temperature plastic deformation behavior of metals and alloys\(^{17}\) which can be generalized as follows;

\[
\dot{\varepsilon} = A \left(\frac{Eb}{kT}\right) \left(\frac{b}{d}\right)^{p} \left(\frac{\sigma}{E}\right)^{n} D_{0}\exp\left(-\frac{Q}{RT}\right)
\]

where \(\dot{\varepsilon}\) is the steady-state strain rate, \(A\) is the dimensionless constant, \(E\) is the elastic modulus (= 2.67 \times (19200 - 8.67) MPa for magnesium\(^{10}\)), \(b\) is the Burgers vector, \(k\) is the Boltzman’s constant, \(T\) is the absolute temperature, \(p\) is the grain size exponent, \(\sigma\) is the flow stress, \(n\) is the stress exponent, \(D_{0}\) is the pre-exponential factor for diffusion, \(Q\) is the activation energy for diffusion, and \(R\) is the gas constants (8.3144 J/mol K). According to eq. (1), the activation energy, \(Q\), can be determined by plotting the logarithmic \(\dot{\varepsilon}_{c}(E_{T}/E_{753})^{n}(T/753)\), which is normalized to temperature of 573 K, against \(1/T\),\(^{10}\) based on the following relation;

\[
Q_{c} = \left[\beta \ln(\dot{\varepsilon}_{c}E_{753}^{-1}T)/\beta (-1/RT)\right]_{T}
\]

Activation energies for plastic flow were computed in the low and intermediate stress regions in Fig. 1 where the data fit for \(n = 3\) and \(n = 5\) respectively. If the rate controlling mechanism in the intermediate stress region is dislocation climb creep, being probable since the data associated with this region show \(n = 5\), \(Q_{c}\) should be equal to \(Q_{L}\), activation energy for lattice self-diffusion in pure magnesium (=135 kJ/mol\(^{18,19}\)). The energy, \(Q_{c}\) (=125 ± 1 kJ/mol), which is computed by measuring the slope of the lines in Fig. 2, is indeed similar to \(Q_{L}\). This value also agrees with that (≈124 kJ/mol) reported on the fine-grained AZ61 Mg alloy (\(d \sim 16\mu m\)) showing \(n = 5\) in the intermediate stress region.\(^{15}\) In the low stress region where data fit for \(n = 3\), on the other hand, VGC can be admitted as the dominant deformation mechanism if the measured activation energy in Fig. 2 is equal to that for inter-diffusion of Al in the Mg lattice\(^{10}\). Indeed, the two values (127 ± 5 kJ/mol vs. 143 ± 10 kJ/mol) are similar. It is noted that these values are similar to that for the Mg-0.8%Al alloy \((Q = 140 ± 10 kJ/mol\(^{11}\)) measured in the stress region showing \(n = 3\).

For analyzing alloying effect on Mg alloy’s creep transition with VGC, other Mg alloys (Mg-0.5, 1, 3%Al\(^{12}\) Mg-3%Al\(^{13}\) Mg-5%Al\(^{13}\) AZ91-20 vol% Al\(_{2}\)O\(_{3}\)\(^{(14)}\)), where \(f\) denotes fibers) are employed. It is reasonable that materials containing below 3 mass%Al used in this work are solid solution because solutes in AZ31 dissolved fully in the Mg matrix.\(^{21}\) Exceptionally, it is noted that AZ91-20 vol% Al\(_{2}\)O\(_{3}\)\(^{(14)}\) was reported to show similar creep behavior with dilute solid-solution of Mg alloy because of its reduced Al’s solubility (near to ∼2 mass% at the aging temperature of 443 K) in Mg matrix resulted from the presence of second phase Mg\(_{17}\)Al\(_{12}\) or Mg\(_{17}\)(Al,Zn)\(_{12}\) intermetallics below 710 K.

3. Results and Discussion

3.1 General creep behaviors (DCC and VGC)

The data associated with \(n = 5\) (intermediate stress region) in Fig. 1 were plotted as \(\dot{\varepsilon}(Eb/kT)^{-1}D_{c}^{-1}\) vs. \(\sigma/E\) in Fig. 3(a) and compared with the various Mg or Mg alloys reported to DCC by other investigators: pure Mg \((d \sim 139\mu m)\),\(^{10}\) Mg-
0.8% Al (d = 417 µm), 11 Mg-0.5%, 1, 3% Al (d ~ 170 µm),12 Mg-3% Al (d = 156 µm),13 Mg-5% Al (d = 136 µm),13 AZ31 (d = 54 µm),13 AZ61 (d ~ 16 µm),15 AZ91-20 vol% Al2O3 (f)14 and AZ91D-cast (d = 300 µm).22 The comparison shows that 1) influence of grain size on creep resistance is not significant, agreeing with the general observation for DCC being independent of grain size15,17 and 2) increasing Al content increases the creep strength slightly (i.e., decreasing strain rate at a given stress). This is directly shown in Fig. 4(a) by measuring the values of A (called as ADCC for n = 5, D0 is D0L = 1 × 10⁻⁴ m²/s¹⁰⁰) in eq. (1) for the materials in Fig. 3(a) as a function of Al content (converted into the mol%). For the alloys, the value of ADCC is in range between 3.0 × 10⁻⁵~1.1 × 10⁻³. For pure Mg, ADCC is 1.22 × 10⁻³.

A phenomenological equation for DCC has been developed by Sherby and coworkers5,23 that is as follows:

$$\dot{\varepsilon} = 7.04 \times 10^9 \left( \frac{Eb}{kT} \right) \left( \frac{\sigma}{E} \right)^5 D0L \exp\left(-\frac{Q}{RT}\right)$$  (3)

The deformation behavior predicted by eq. (3) is shown as a solid line in Fig. 3(a). It is revealed that the pure Mg and Mg alloys are notably stronger than the predicted. It should be reminded, however that eq. (3) was developed for coarse grained, non-textured and high stacking fault energy metals having FCC and BCC structures. Mg with HCP structure has lower stacking fault energy (i.e., γMg ~ 0.04 J/m²⁴) than typical metals with FCC or BCC structure (for example, γAl is ~0.20 J/m²⁴). Besides, Mg has a limited number of slip systems. With these strengthening factors causing more restriction on dislocation movement, higher level of creep resistance by adding Al is expected in magnesium.

Weertman has developed a theoretical model for DCC given as below:4,10

$$\dot{\varepsilon} = \alpha' \left( \frac{\Omega}{b\sqrt{M}M'} \right) \left( \frac{Eb}{kT} \right) \left( \frac{\sigma}{E} \right)^{4.5} D0L \exp\left(-\frac{Q}{RT}\right)$$  (4)

where α' is the constant (α' = 31 × α, α (0.015~0.331⁰) is assumed to be 0.1 in this work), Ω is the atomic volume (2.38 × 10⁻¹⁰ m³¹⁰) and M is the density of Frank-Read sources that can be determined from the relation of M⁰.⁵ = 1/1.9N⁰.⁷⁵, where N is the density of dislocations.⁶ Two broken lines in Fig. 3(a) denote two extremes of n = 0, 2, 4, 8, 16, which are ADCC for various Mg alloys.

![Figure 3](image-url) Normalized strain rate vs. normalized stress predicted by models and measured experimentally for the dominant deformation mechanisms of (a) DCC and (b) VGC.

![Figure 4](image-url) (a) ADCC and (b) AVGC for various Mg alloys.
by solute-solvent size difference will promote the increase of creep strength.

The increase of creep resistance by addition of Al in the region associated DCC may be linked with decrease in stacking fault energy (γ) with increase of Al in Mg. When the γ is low, dislocations are widely split with a fault between them. Such split dislocations find it difficult to climb, hence one would expect to observe a slower rate of creep for such materials than high γ materials. Experimental data for the stacking fault energies in Mg-Al or Mg-Al-Zn alloy are not available in literatures. There has been, however, a recent attempt to compute γ_{alloy} in several solid solution Mg alloys using the first-principles calculations. According to it, the γ_{alloy} decreases from 0.040 to 0.020 J/m² for Al and 0.040 to 0.035 J/m² for Zn when each solute is added by 3.57%. 27) However, the Mg-0.5%Al alloy and AZ31 alloy in Mg matrix is available only at 583 K in Mg-5\%-Zn-0.6\%Zr alloy, 28) we simulate deformation rate at 583 K for Mg in AZ31 alloy: \( \dot{\varepsilon}\text{Al in Mg} = 1.85 \times 10^{-10} \text{m}^2/\text{s} \) and \( \dot{\varepsilon}\text{Zn in Mg} = 1.22 \times 10^{-15} \text{m}^2/\text{s} \). The calculated value of 6.0 \times 10^{-4} (Ds/e²C₀) is 5.70 \times 10^{-12} \text{m}^2/\text{s} and 2.53 \times 10^{-16} \text{m}^2/\text{s} for Mg-1\%Zn and Mg-3\%Al, respectively. As the value for Mg-1\%Zn is higher, deformation rate for VGC at 583 K at a fixed flow stress is expected to be dominated by Al atmosphere rather than Zn atmosphere. Reduced Al content in matrix due to a possibility of compound formation between Al and Zn or increase in Al’s chemical diffusivity due to change of thermodynamic factor of Al in the presence of Zn may be the reason for the increase in deformation rate by addition of Zn. However, Zn effect on VGC in the Mg-Al-Zn system is still not clear because of lack of data, hence, systematic study is required to confirm above speculation and such study is now in progress.

3.2 Transition of deformation mechanism in Mg alloys

To observe the transition behavior of deformation mechanism in the pure Mg and Mg alloys, the data for pure Mg, Mg-0.8\%Al alloy and AZ31 alloy in the temperature range (523~763 K) are plotted as \( \log \dot{\varepsilon} \) vs. \( \log \sigma/E \) in Fig. 5. The transition from VGC to DCC as shown in various solid solution alloys, 9,30~34 occurred in the Mg-0.8\%Al and AZ31 alloys with increase in stress level shown in Figs. 1 and 5 may be due to the breakaway of mobile dislocations from their solute atmospheres. The critical breaking stress (breakaway stress), \( \sigma_b \), is expressed as according to Vagarali and Langdon.11)

\[
\dot{\varepsilon} = A_{\text{VGC}} \left( \frac{E_b}{kT} \right) \frac{\sigma^3}{E} D_\text{OS} \exp\left( -\frac{Q}{RT} \right)
\]

(5)

where \( D_\text{OS} = 1.2 \times 10^{-3} \) (Al’s chemical interdiffusivity in Mg matrix, m²/s/Å³), and \( Q = Q_\text{Al in Mg} = 143 \pm 10 \text{kJ/mol} \). Values of \( A_{\text{VGC}} \) for the solid solution Mg alloys are within a range between 3.2 \times 10^{-4} to 1.2 \times 10^{-1} for Mg-Al alloys and 4.9 \times 10^{-2} \sim 8.7 \times 10^{-1} for Mg-Al-Zn alloys. The alloying effect on \( A_{\text{VGC}} \) (i.e., creep resistance) is shown in Fig. 4(b). It is apparent that the \( A_{\text{VGC}} \) value for the Mg-Al alloy decreases with increase in Al and similar tendency keeps for Mg-Al-Zn alloys. In Fig. 3(b), the deformation rate of Mg at different concentration of Al (and Zn) is predicted by Weertman’s theoretical model for VGC or using the following equation;

\[
\dot{\varepsilon} = 7.94 \times 10^{38} \pi (1 - v) \frac{1}{e^2 C_0} \left( \frac{E_b}{kT} \right)^{-1} \left( \frac{\sigma}{E} \right)^3 D_s
\]

(6)

where \( v \) is Poisson’s ratio (usually 0.35 in Mg), \( e \) is the solute solvent size difference, \( C_0 \) is the solute concentration, \( D_s \) is solute’s interdiffusivity in Mg matrix. eq. (6) can be converted to the form of eq. (1) by letting \( E_b/kT = 1.64 \times 10^3 \text{m}^2/\text{s} \) which was determined from the average value of (1.64 \pm 0.47) \times 10^3 \text{m}^2/\text{s} at temperature range of 423 to 698 K in the explicit form of

\[
\dot{\varepsilon} = 6.00 \times 10^{-4} \frac{1}{e^2 C_0} \left( \frac{E_b}{kT} \right) \left( \frac{\sigma}{E} \right)^3 \frac{D_s}{D_\text{OS}} \exp\left( -\frac{Q}{RT} \right)
\]

(7)

Comparison of the prediction with the experimental results indicates that the theory overestimates the deformation rate. Nevertheless, it is admitted that the equation can correctly predict the trend of alloying effect since it shows \( A_{\text{VGC}} \propto 1/C_0 \). There is, however, a gap of deformation rate as wide as about 10 times or more between Mg-Al and Mg-Al-Zn alloys in VGC.

As an account for above divergence between Mg-Al and Mg-Al-Zn alloys, chemical interdiffusivity of each solute is considered. If a mobile dislocation drags solute atmospheres of Al or Zn at the same time, VGC will be determined by slower rate between two simultaneous cases. By the eq. (7), when VGC occurs, the rate of mobile dislocation, which drags solute atmosphere, is determined by a combined term of \( 6.0 \times 10^{-4}(D_s/e^2C_0) \), m²/s. The solute-solute size difference (e) were \(-0.1373 \text{ for Mg-Al alloy and } -0.2000 \text{ for Mg-Zn alloy}.26) \text{ As Zn’s chemical interdiffusivity in Mg matrix is available only at 583 K in Mg-5\%-Zn-0.6\%Zr alloy,28) we simulate deformation rate at 583 K for VGC in AZ31 alloy: } D_\text{Al in Mg} = 1.85 \times 10^{-10} \text{m}^2/\text{s} \text{ and } D_\text{Zn in Mg} = 1.22 \times 10^{-15} \text{m}^2/\text{s}. The calculated value of 6.0 \times 10^{-4} (D_s/e^2C_0) is 5.70 \times 10^{-12} \text{m}^2/\text{s} and 2.53 \times 10^{-16} \text{m}^2/\text{s} for Mg-1\%Zn and Mg-3\%Al, respectively. As the value for Mg-1\%Zn is higher, deformation rate for VGC at 583 K at a fixed flow stress is expected to be dominated by Al atmosphere rather than Zn atmosphere. Reduced Al content in matrix due to a possibility of compound formation between Al and Zn or increase in Al’s chemical diffusivity due to change of thermodynamic factor of Al in the presence of Zn may be the reason for the increase in deformation rate by addition of Zn. However, Zn effect on VGC in the Mg-Al-Zn system is still not clear because of lack of data, hence, systematic study is required to confirm above speculation and such study is now in progress.
where $\sigma_b$ is the binding energy between solute and dislocations\(^7\) which is defined as

$$W_M = \frac{W_M^2 C_0}{5b^3/kT}$$

where $\Delta V_s$ is the difference in volume between the solute and solvent atoms ($8.3 \times 10^{-30}$ m\(^3\) for Mg-Al and $11.3 \times 10^{-30}$ m\(^3\) for Mg-Zn\(^28\)). When $\sigma > \sigma_b$, the solute atmosphere no longer retards the movement of dislocations and as the result a transition from VGC to DCC occurs. With eqs. (8) and (9), $\sigma_b$ values for the Mg-Al and Mg-Zn alloys could be computed. In Fig. 6(a), experimentally determined breakaway stresses of AZ31 and Mg-Al alloys measured\(^a\) (with VGC, surrounded by black dot-line set) and Mg alloy\(^{15}\) (with VGC, surrounded by gray solid line-set). Loaded values were experimentally determined for Mg-Xmass%Al ($X = 0.5, 0.8, 1$ and $3$) and AZ31 and theoretically predicted for the Mg-1 mass%Zn and Mg-1.3 mass%Al alloys.

that the experimentally determined breakaway stresses of the highly alloyed solid solutions are higher than those of the dilute alloy and the theory predicts the trend. When theoretically predicted $\sigma_b$ for Mg-3%Al and Mg-1%Zn compared with experimental results, real data for AZ31\(^{11,15,16}\) are in line with Mg-3%Al’s line. This support above speculation that breakaway will occur from VGC Mg-Al to DCC rather than VGC Mg-Zn to DCC. Lee and Langdon reported a discrepancy between theoretical equation (eq. (8)) and AZ91 alloy since the reduced solubility of Al by the existence of second phase or intermetallics.\(^14\) Indeed, experimental breakaway stress of AZ91-20 vol%Al\(_2O_3\)\((f)\) was ~14 MPa close to ~19.7 MPa (Mg-2 mass%Al) rather than ~88 MPa (Mg-9 mass%Al) at 573 K by eq. (8). It is noted that $\sigma_b$ depends on the actual composition (~2 mol% after 24 h aging at 443 K) rather than the initial composition (AZ91). Nevertheless, it is noted, however, that the theoretical breakaway stress deviates from the data for the Mg-Al alloys with Al less than 3 mass% quantitatively.

In Fig. 6(b), experimentally determined\(^{11,13,15,16}\) and theoretically predicted $\sigma_b$ values of the Mg-1 mass%Zn and Mg-Xmass%Al alloys ($X = 1$, 2, and 3) are plotted in deformation mechanism maps which are constructed at 623 K; one for pure Mg (free from VGC) and the other\(^{15}\) for Mg alloy. In Fig. 6(b), VGC’s area (surrounded by solid line-set) vanishes and will be possessed by Harper-Dorn creep, grain boundary sliding and DCC when no VGC is assumed (gray dot-line set). Experimental data (points 1~6) and analytical lines (underlined points 7~9, which were calculated by eqs. (8) and (9) for Mg-1%Zn, Mg-1%Al, Mg-3%Al, Mg-3%Al) falls close around the line between VGC and DCC in the map. They keep the tendency discussed above paragraph, i.e., breakaway stress increases with Al contents, and moreover, shows no dependency of grain size. Hence, this map can be appreciated because it helps to understand the transition from VGC to DCC and to show no dependency of grain size on breakaway when combining with Friedel’s model.

Another transition from $n = 5$ to $n \geq 7$ observed at high stress level is believed to be linked with power law breakdown (PLB) that frequently occurs in many metallic alloys at high stresses. Kim et al.\(^{15}\) reported that this transition occurs in Mg when $D_P$ controlled DCC starts to dominate the plastic flow. The transition occurs $\sigma/E = 2-3 \times 10^{-3}$ according to the deformation mechanism map at 573 K and 673 K and indeed these values are close to the experimentally determined values of 1.3~1.6 $\times 10^{-3}$ for DCC to PLB at 573 K and 648 K respectively.

4. Summary

(1) In DDC, Mg or Mg alloys showed higher creep strength than that of typical alloys may result from the low stacking fault energy ($\gamma$) of magnesium and/or solid solution strengthening. Moreover, Mg alloys containing higher Al content have higher creep strength than low content Mg alloys because $\gamma$ in Mg decreases by adding solute atoms. However, Zn in Mg alloys seems no remarkable effect in DCC with its relatively lower effect on $\gamma$ in Mg and low content.
(2) In VGC, creep resistance increased with Al content in both Mg-Al and Mg-Al-Zn alloys. Addition of Zn to Mg-Al is analyzed to increase the creep rate by more than one order. Decreased Al content in matrix by compound formation between Al and Zn or increase in Al’s chemical diffusivity in the presence of Zn may be the reason.

(3) Transition from VGC to DCC in the coarse grained Mg alloys depends on the content of Al contents in either Mg-Al or Mg-Al-Zn alloys. Breakaway stress in Mg alloys shows the dependency on Al content, increasing with Al content. Theoretical breakaway stress can predict the transition from VGC to DCC reasonably well for highly alloyed solid solution (Mg-3%Al and AZ31), but deviates in a quantitative prediction for dilute Mg-Al alloys.

(4) Deformation mechanism map for Mg alloys can be used to predict the occurrence of breakaway behavior when combined with Friedel’s model, then showed no dependency of grain size on transition from VGC to DCC.

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