Strain Enhanced Precipitate Coarsening during Creep of a Commercial Magnesium Alloy AZ80

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Growth of precipitates during creep was investigated of a commercial magnesium alloy AZ80. TEM observations in this study confirmed that the cube of the average particle size changed roughly with creep time at grip parts and the growth rate was greater at the gauge part than at the grip part of crept specimens. The cube of the average particle size changed roughly with creep time at grip parts. This suggests that the coarsening of particles at grip part obeyed the Ostwald ripening theory in which the growth rate was assumed to be controlled by the lattice diffusion. In order to make clear the factors affecting the growth rate of precipitates at gauge part, interrupted creep tests were carried out. The present analysis of the experimental results showed that the effect of strain on the particle growth was more important than that of stress. The influence of concurrent straining on the particle growth is also discussed in terms of traveling dislocations with solute atmospheres.

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Keywords: magnesium alloy, creep, Ostwald ripening, particle growth, precipitate, solute atmosphere

1. Introduction

Precipitates embedded in matrix can be an obstacle to dislocation motion, so that they contribute greatly to the strength of metals and alloys. In fact, a large number of commercial heat resistant materials rely for their high-temperature strength on the presence of finely dispersed precipitates. Precipitation-hardening is, however, deteriorated due to the growth of precipitates progressing at elevated temperatures. Such growth of precipitates inevitably leads to the decrease of threshold stress to dislocation motion, and consequently the loss of high temperature strength occurs. For these reasons, many investigators have paid attention to the Ostwald ripening which appears at the final stage of precipitation processes.

During the last decade, various approaches have been attempted to improve the classical Ostwald ripening theory (LSW theory), however, the power law of the LSW theory remains nearly unchanged as many experimental facts show. Among the previous studies on this subject, the work done by Spigarelli et al. seems to be worthwhile to consider closely. They showed that the precipitates grew larger at the gauge parts than at the grip parts of crept specimens and the larger the applied stress was, the larger the particle size was. Actually, they attributed this interesting phenomenon to the effect of creep stress, however, this can also be regarded as the effect of strain because a larger applied stress produces a larger creep strain at a given time. In fact, the reanalysis of the results obtained by Spigarelli et al. showed that the enhanced growth of particles during creep could be accounted for by the difference in creep strain. In order to examine whether such enhanced growth was unique to some ferrous alloys or not, the present author has studied the effect of creep deformation on the precipitate growth. The previous work done by the present authors confirmed the enhanced coarsening of precipitates in an Al–Cu alloy.

In the present study, examination was carried out to make clear which factor, stress or strain is more important for the acceleration of the growth of precipitates in a commercial magnesium alloy of AZ80. Experimental results were also discussed on the Cohen’s criterion, which determined the minimum strain rate necessary to observe the enhanced self-diffusivity during plastic deformation.

2. Experimental Procedure

A commercial magnesium alloy (AZ80) manufactured by Timminco Inc. was used in the present study. According to the prescription of the nominal composition of the alloy, the magnesium base AZ80 alloy contains approximately 8 mass% aluminum, and 1 mass% zinc, the chemical composition of minor elements of the as-received material is given in Table 1. An as-received specimen was already treated with the T5 heat-treatment at 443 K for 16 h. Taking the condition of the treatment into account, it was expected that only the γ phase precipitates were formed in the alloy and they were in the stage of Ostwald ripening.

Flat creep specimens were machined from the as-received round bar. The dimensions of a creep specimen are shown in Fig. 1. The cross sectional area of the gauge section was 10 mm² and the area of the grip section was four times larger. Therefore, the stress at grip section was one-forth of the stress at gauge section. Since the stress exponent of the present alloy is approximately 4.6, both strain and strain rate at the grip part are negligible.

Conventional creep tests were carried out at 433 K. Before the application of creep stress, the specimen was held at

<table>
<thead>
<tr>
<th>Table 1 Chemical composition of AZ80.</th>
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<tr>
<td>Al</td>
</tr>
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<tr>
<td>8.0</td>
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433 K for 24 h to stabilize the temperature of the assembly composed of specimen, pull rods and extensometer. The initial creep stress at a gauge section was 37.0, 49.0 or 58.0 MPa, and each sequence of the creep test was interrupted when the strain reached 0.1. In addition to these specimens, the other set of specimens which were crept up to a strain of 0.05 at a stress of 37.0 MPa was also examined. The average strain rates during creep were in the range of $5 \times 10^{-8}$ to $2.3 \times 10^{-7}/s^{-1}$.

To examine the morphology of precipitates, some SEM observations were carried out of specimens which were solution treated at 713 K and then aged at 433 K for 48 h. To avoid the effect of surface oxidation during aging, the surface layer of specimens was removed mechanically and then electro-polished. For the measurements of precipitate size, TEM observation was conducted before and after creep.

3. Results

3.1 Microstructure

An SEM micrograph is shown in Fig. 2, where the colony of acicular precipitates (indicated by the letter, “A”) is seen in the vicinity of the grain boundary of a round grain marked by the letter, “B”. The colony of the precipitates was caused as a result of a boundary reaction occurring in this alloy. Figure 2(b) is an SEM micrograph of the grain B taken at higher magnification. A large number of fine and globular precipitates are observed in the grain B. The energy dispersive X-ray (EDX) analysis revealed that both of the precipitates have a same chemical composition corresponding to $\text{Mg}_{17}\text{Al}_{12}$. Electron diffraction analysis confirmed that its unit cell is $\alpha$-Mn type cubic structure. The present results agreed to those in previous works.\(^{16,17}\)

3.2 Measurements of the particle sizes

To consider the precipitation behavior in the alloy, the changes in the size and density of the precipitates are essentially important to obtain in this study. We estimated the size and the density of the precipitates from bright field TEM images. As shown in Fig. 2(a), acicular precipitates due to a boundary reaction are found, but the area with the precipitates was limited. Therefore, only the particles isolated inside grains were measured in the present study. Although most of the particles showed granular shapes, they were approximated by ellipses, and the lengths of both the major and minor axes of each precipitate were measured in the TEM images to determine the diameter of a circle having an equivalent area to the ellipse. The relative frequency of the particle diameter is shown in Fig. 3, in which approximately three-hundred and sixty particles were taken into account in the as-received material. Almost all particles are less than 100 nm in size, and the mean diameter was 66 nm. In order to examine the influence of creep on the particle growth, the equivalent particle diameter was measured for the precipitates appearing at the gauge and grip parts of the crept specimens.

Figures 4(a) and (b) show TEM images of the microstructures formed at the gauge and grip parts, after the creep-test at 49 MPa for $1.6 \times 10^6 \text{s}$, respectively. According to Celotto,\(^{16}\) the shape of precipitates in this alloy is plates or laths at the early stage of precipitation, while they grow to granular ones for further aging. Granular shape of particles in grip and gauge parts as seen in Fig. 4(a) and Fig. 4(b)

![Fig. 1 Dimensions of a creep specimen.](image)

![Fig. 2 SEM images of an AZ80 alloy. (a) A round grain is indicated by B, and the colony of acicular precipitates, which were formed by a grain boundary reaction is indicated by A. (b) A SEM image showing the globular precipitates within the B grain with higher magnification.](image)
suggests that the precipitation is almost complete and the Ostwald ripening process still begins to occur. Some particles exhibit black image contrast but others do not. The difference in contrast may suggest that particles do not have a strict orientation relationship to the matrix, but they are randomly distributed in matrices. In the present study, the difference of image contrast was not taken into account for the determination of equivalent diameters. Comparing Figs. 4(a) and (b), we noticed that the precipitate particles formed at the gauge parts are explicitly larger than those at grip parts. The histograms showing the distribution of equivalent diameters are shown in Figs. 5(a) and (b), and the mean diameter of the precipitates is listed in Table 2 for various creep conditions. The present result was drawn from the measurement of about 100 to 150 particles collected in about 20 different observation areas, for each creep condition. Figures 5(a) and (b) shows that a half of the small particles less than 50 nm in the grip part disappeared in the gauge part. This means that the relatively large particles grow with the absorbing small precipitates in the vicinity.

Fig. 3  The distribution of the equivalent particle diameters of the as-received AZ80 alloy.

Fig. 4  TEM micrographs showing the precipitates at a gauge part (a) and a grip part (b) of a crept specimen tested under a stress of 45.8 MPa at 433 K for $1.4 \times 10^6$ s.

Fig. 5  The histograms showing the distribution of equivalent diameters at a gauge part (a) and a grip part (b) of crept specimen.
According to the classical Ostwald ripening theory,\textsuperscript{(11-13)} Equation (1) is applicable under the condition that the total fraction of precipitates in volume is constant, and its coarsening rate is controlled by the lattice diffusion of the solute.

\[ d^3 - d_0^3 = \left( \frac{64 \gamma C_\infty D_L}{9RT} \right) t, \]  

(1)

where \( d \) and \( d_0 \) are the precipitate diameter at time \( t \) and \( t = 0 \), respectively, \( \gamma \) is the interfacial energy, \( C_\infty \) is the equilibrium concentration of solute at the interface between particle with infinite diameter and matrix, \( D_L \) is the lattice diffusivity of solute, \( R \) the universal gas constant, and \( T \) is the testing temperature.

In the present creep tests, a specimen was held at testing temperature for 24 h before the application of stress. So, taking \( t_1 \) for the time spent before the application of stress, and \( t_2 \) for creep time, one obtains eqs. (2) and (3).

\[ d_1^3 - d_0^3 = K t_1, \]  

(2)

and

\[ d^3 - d_0^3 = K (t_1 + t_2), \]  

(3)

where \( d_1 \) and \( d \) are particle size at \( t = t_1 \) (aging time spent at testing temperature before creep) and \( t = t_2 \) (aging time during creep), respectively, and \( K \) is given by eq. (4).

\[ K = \left( \frac{64 \gamma C_\infty D_L}{9RT} \right). \]  

(4)

With eqs. (2) and (3), one obtains the following equation.

\[ d^3 - d_1^3 = K t_2. \]  

(5)

Based on the above equations, the relationship between the cubic of the average precipitate diameter and creep time is shown in Fig. 6. The numerals attached to each plot in Fig. 6 correspond to each creep condition given in Table 2. The asterisk ‘*’ at the shoulder of the numeral designates the data of gauge parts, while plots of 1, 2, 3 and 4 without asterisk show the data of the grip part. So the strain corresponding to grip part is substantially negligible. On the other hand, plots of 1*, 2*, 3* and 4* indicate the data of gauge parts and the strain of 1*, 3* and 4* is 0.1 while that of 2* is 0.05. It should be noted here that the applied stress of 9 MPa is common to plots 2 and 4, while the stress of 37 MPa is common to plots 2* and 4*.

![Fig. 6 Relationship between the cubic of the average precipitate diameter and creep time for gauge and grip parts. Time is taken from the onset of creep deformation.](image)

The present result shown in Fig. 6 revealed an interesting tendency of the coarsening of precipitates in the AZ80. That is, the cube of the average precipitate diameter at the grip part can be expressed by a linear relation of creep time, in spite of a distribution of the datum points. This suggests that the growth of precipitates at the grip part can be formulated by the mechanism based on the Ostwald ripening theory controlled by the lattice diffusion of solutes. In contrast, the growth rate at the gauge part is obviously faster than that at grip part, namely, the coarsening is enhanced at the gauge part.

### 4. Discussion

#### 4.1 Factors affecting the particle growth

The results shown in the previous section demonstrate clearly that the coarsening rate of precipitates is higher at gauge parts than at grip parts. This suggests that the diffusivity of solutes is enhanced at gauge parts. Since a larger applied stress produces a larger strain at a given creep time, it is difficult to conclude readily which factor, stress or strain, is more important for the enhanced coarsening. However, when coarsening is assisted solely by stress or strain, one can draw the conclusion which factor is more dominant on the experimental basis. Hereinafter, two extreme cases where only the stress or strain can assist the diffusivity of solutes will be examined.

#### 4.2 Stress dependent coarsening

As we have seen so far, the growth rate of particle was obviously faster at gauge parts than at grip parts. This fact suggests that the solute diffusivity is enhanced at gauge parts. Now, let us assume that the effective diffusivity of solutes is stress dependent. In such a case, the effective diffusivity is equal to lattice diffusivity when stress is absent. Using eq. (1), the cube of precipitate diameter can be expressed as below for the case of 2*, 4* and 4*.

\[ d_2^3 - d_0^3 = A \cdot (D_{eff}|_{\sigma=0}) \cdot t_2. \]  

(6)

\[ d_4^3 - d_0^3 = A \cdot (D_{eff}|_{\sigma=37}) \cdot t_2. \]  

(6')

Table 2  Creep conditions and precipitate size.

<table>
<thead>
<tr>
<th>Stress ((\sigma/\text{Mpa}))</th>
<th>Creep time ((t/\text{s}))</th>
<th>Strain ((\varepsilon))</th>
<th>Mean diameter of precipitates ((d/\text{nm}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>14</td>
<td>4.4 ( \times ) 10(^3)</td>
<td>81</td>
</tr>
<tr>
<td>2* grip</td>
<td>9</td>
<td>1.0 ( \times ) 10(^6)</td>
<td>139</td>
</tr>
<tr>
<td>3</td>
<td>12</td>
<td>1.0 ( \times ) 10(^6)</td>
<td>0</td>
</tr>
<tr>
<td>4</td>
<td>9</td>
<td>1.6 ( \times ) 10(^6)</td>
<td>149</td>
</tr>
<tr>
<td>1* gauge</td>
<td>58</td>
<td>4.4 ( \times ) 10(^5)</td>
<td>0.1</td>
</tr>
<tr>
<td>2*</td>
<td>37</td>
<td>1.0 ( \times ) 10(^6)</td>
<td>0.05</td>
</tr>
<tr>
<td>3*</td>
<td>49</td>
<td>1.0 ( \times ) 10(^6)</td>
<td>0.1</td>
</tr>
<tr>
<td>4*</td>
<td>37</td>
<td>1.6 ( \times ) 10(^6)</td>
<td>0.1</td>
</tr>
</tbody>
</table>

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Strain Enhanced Precipitate Coarsening during Creep of a Commercial Magnesium Alloy AZ80 1101
\[ d_4^2 - d_0^2 = A \cdot (D_{eff}|_{\sigma=0}) \cdot t_4. \]  
\[ d_4^3 - d_0^3 = A \cdot (D_{eff}|_{\sigma=37}) \cdot t_4. \]

where \( d_i \) is the mean diameter corresponding to each datum plot, \( A \) is a constant with a magnitude of \((647C_{ec}/9RT))\), and \((D_{eff}|_{\sigma})\) is the effective diffusivity depending on stress, \( \sigma \). Eliminating \( d_3 \) which is common to any cases, one obtains eqs. (8) and (8').

\[ d_2^3 - d_2^3 = A((D_{eff}|_{\sigma=37}) - (D_{eff}|_{\sigma=0})t_2. \]  
\[ d_4^3 - d_4^3 = A((D_{eff}|_{\sigma=37}) - (D_{eff}|_{\sigma=0})t_4. \]

Further eliminating the common terms in eqs. (8) and (8'), then eq. (9) results.

\[ \frac{d_2^3 - d_2^3}{d_4^3 - d_4^3} = \frac{t_2}{t_4}. \]

Equation (9) predicts that the difference between \( d_2 \) at gauge parts and \( d_2 \) at grip parts is proportional to creep time when creep tests are conducted at the same stress. Using the experimental data, the magnitudes of the right and left hand sides of eq. (9) are calculated to be 0.6 and 0.3, respectively. The disagreement between them demonstrates that the hypothesis to deduce eq. (9), namely, the effective diffusivity is stress dependent, is not valid in this case.

### 4.3 Strain dependent coarsening

Since the assumption that the solute diffusivity is stress dependent was discarded in the previous section, an alternative assumption, namely the solute diffusivity is strain dependent, will be examined. On this assumption, the effective solute diffusivity is equal to lattice diffusivity when strain is absent. So the coarsening behavior for plot numerals of 3, 3', 4 and 4' can be expressed as below using eq. (1):

\[ d_3^3 - d_0^3 = A \cdot (D_{eff}|_{\sigma=0}) \cdot t_2. \]  
\[ d_3^4 - d_0^4 = A \cdot (D_{eff}|_{\sigma=0}) \cdot t_2. \]  
\[ d_4^3 - d_0^3 = A \cdot (D_{eff}|_{\sigma=0}) \cdot t_4. \]  
\[ d_4^4 - d_0^4 = A \cdot (D_{eff}|_{\sigma=0}) \cdot t_4. \]

where \((D_{eff}|_{\sigma})\) is the effective diffusivity depending on strain. Using eqs. (10), (10'), (11) and (11'), the difference between the cubes of precipitate diameter at gauge and grip parts is given as below.

\[ d_3^3 - d_3^3 = A((D_{eff}|_{\sigma=0.1}) - (D_{eff}|_{\sigma=0})t_2. \]  
\[ d_4^3 - d_4^3 = A((D_{eff}|_{\sigma=0.1}) - (D_{eff}|_{\sigma=0})t_4. \]

Comparison of eqs. (12) and (12') leads to the following equation.

\[ \frac{d_3^3 - d_3^3}{d_4^3 - d_4^3} = \frac{t_2}{t_4}. \]

Using the experimental data, the magnitudes of the right and left hand sides of eq. (13) were calculated to be 0.6 and 0.7, respectively. This excellent agreement between them suggests strongly that the enhanced coarsening at gauge parts is unable to be ascribed to stress but strain. Based on this conclusive remark, it is easy to explain the fact that the growth rate of 3' is higher than that of 2' because creep strain (\( \varepsilon \approx 0.1 \)) of 3' is larger than the strain (\( \varepsilon \approx 0.05 \)) of 2', when compared at the same creep duration.

### 4.4 Comparison with Cohen's criterion

Cohen et al.\(^{15}\) have demonstrated experimentally that the ratio of self-diffusivity during plastic deformation, \( D' \) (dynamic self-diffusivity) with static self-diffusivity, \( D \) is expressed by eqs. (14) and (15).

\[ \frac{D'}{D} = 1 + B(T)\dot{\varepsilon}, \]

and

\[ B(T) = B_0 \exp(B'/RT), \]

where \( B(T) \) is a measure of the diffusion sensitivity to plastic strain rate, and \( B_0 \) and \( B' \) are positive constants. According to their study, \( B(T) \) is independent of strain, strain rate and temperature over a wide range of experimental conditions.\(^{15}\) On the contrary to this, the growth rate of precipitates obviously depends on plastic strain in the present study. This suggests that the diffusion coefficient is dependent upon plastic strain.

Cohen et al.\(^{15}\) have developed a model to account for the increase in diffusion during plastic deformation by combining the empirical eqs. (14) and (15) with the average diffusivity given by Hart.\(^{16}\)

\[ D' = D + D_A \rho \]

where \( D' \) is the normal diffusivity in the matrix, \( D_A \) is the diffusivity in dislocation pipes, \( A \) is the diffusion cross section of dislocation pipes, and \( \rho \) is the density of dislocations randomly distributed in matrix. Equation (16) should hold for a random-walk process under a condition where dislocation density is high enough and diffusing atom can enter and leave a dislocation core several times during a given measurement.\(^{19}\) If dislocations are traveling under the above conditions, the conditions required for eq. (16) can be readily achieved even if their density is lower than that of stationary dislocations. This is because the moving dislocations are able to visit and leave the diffusing species. By substituting the dislocation density expressed by the Ornstein’s equation, \( \dot{\varepsilon} = \phi b \rho \) (\( b \) is the Burgers vector, \( \phi \) is the conversion factor from shear strain to tensile strain, \( \rho \) the dislocation density, and \( \dot{\varepsilon} \) the dislocation velocity) for \( \rho \) in the second term of eq. (16), the dynamic diffusivity, \( D' \) can be expressed as a function of strain rate.

Cohen et al.\(^{15}\) have deduced the following inequality (17) by making use of empirical eqs. (14) and (15). This inequality provides the minimum strain rate necessary to produce dynamic diffusivity which is two times greater than the static diffusivity. In addition to this, they have demonstrated that the boundary given by the inequality (17) separates nicely the enhanced diffusion data from the data which does not show the diffusion enhancement.\(^{15}\)

\[ T/T_m \ll 4.9/(2.2 - \log \dot{\varepsilon}). \]
strain rate in the present study is from $5 \times 10^{-8}$ to $2.3 \times 10^{-7}$ s$^{-1}$. Therefore, it is somewhat difficult to ascribe the present diffusivity enhancement to the Cohen’s mechanism because the strain rate calculated from the inequality is close to the strain rate range in the present study. Actually, the strain rate range fails to support the Cohen’s mechanism firmly, however, there are another possibilities which are favorable to the Cohen’s mechanism. First, Equations (14) and (15) were derived under steady state conditions, so if this conditions were violated and dislocation density increases during deformation, the enhancement will begin to depend on strain as well as strain rate. However, the increase in stress during creep was at least 6 MPa i.e., about 10% in the present case, so it is difficult to explain the strain dependence of diffusivity from the increase in dislocation density. Second factor which should be taken into account is the elastic interaction existing between a dislocation and a solute. Actually, eqs. (14) and (15) were derived from the experiments on tracer diffusion, but they are undoubtedly valid for the chemical interdiffusion as Cohen has already mentioned.\(^{15}\) If Equations (14) and (15) are applied to the chemical interdiffusion, the diffusion enhancement is expected to be more pronounced because diffusion species will spend a large fraction of its time within the high-diffusivity path, i.e., $B^*$ in eq. (15) should increase by the magnitude of the binding energy between a dislocation and a solute.\(^{15}\) In the case of AZ80, since the size misfit of an aluminum atom to a solvent atom is relatively large (size misfit is about 10%\(^{20}\)), the elastic interaction is large. In such case, even if the strain rate subject to a creep specimen is smaller than that expected from eq. (17), it might be possible that the enhanced diffusion becomes larger than $2D$. However, it is important to notice that the enhanced particle coarsening is obviously strain dependent notwithstanding the diffusivity enhancement pertaining to the Cohen’s mechanism is independent of strain. Thus, it is probably quite difficult to account for this outstanding difference between the two from the Cohen’s mechanism.

4.5 Scavenging effect

As mentioned in the previous section, the enhanced diffusivity observed in the present study seems different from the dislocation-assisted diffusivity proposed by Cohen et al.\(^{15}\) because the present case depends on plastic strain while the Cohen’s mechanism do not. Thus it is necessary to direct our attention to an alternative mechanism. One of the probable mechanisms is the scavenging effect of dislocations.\(^{21,22}\) As mentioned already, an aluminum atom has a noticeable size misfit to the solvent atom of AZ80 alloy, so that a solute atmosphere is formed around a dislocation due to an elastic interaction.\(^{24}\) Since the drift velocity of solutes given by the Einstein’s equation is higher than the dislocation velocity at least within the solute atmosphere, the solute atmosphere lag behind dislocations. In other words, the moving dislocations drag their solute atmospheres when they continue to move, however, the shape and size of solute atmospheres as well as the solute distribution around the dislocations remain unchanged so long as the dislocation velocity is kept constant, namely at steady state deformation.\(^{25}\) Now let us imagine the case where dislocations assist the diffusivity of solutes in addition to lattice diffusion. In such a case the size of solute atmosphere is expected to shrink and/or its solute concentration should decrease because solute flow occurs from the dislocations to growing precipitates. However, the solute atmosphere remains unchanged actually because dislocations collect solutes from the matrix during their journey. During Ostwald ripening of precipitates, smaller precipitates dissolve into matrix and larger precipitates grow to reduce the total interfacial energy stored in the precipitate-matrix interface. Consequently, the concentration of matrix is kept constant.

In this process, traveling dislocations collect solutes from matrix and assist the solute transport by providing the moving high-diffusivity paths. Therefore, we will call this “the scavenging effect” of a dislocation hereafter.

The concept of scavenging effect is depicted in Fig. 7 schematically, where $L$ is the effective diameter of a solute atmosphere, $s$ is the displacement of a traveling dislocation, and $C_0$ is the equilibrium solute concentration in matrix. When the self-diffusivity is assisted by traveling dislocations, the apparent cross section of a dislocation pipe becomes larger than that of stationary dislocations. In analogy with this, the cross sectional area in a solid solution becomes much larger than that of a deformation-induced self-diffusivity. This is because there is the attractive interaction between a solute and a dislocation. Here we assume that the effective cross section of the scavenging effect is equal to $L \cdot s \cdot C_0$. On the other hand, there is a relationship between $s$ and plastic strain, $\varepsilon$ as expressed by eq. (18).

$$\varepsilon = \phi p b s. \quad (18)$$

$$D_{\text{eff}} = D_L + (L C_0 / b \phi) D_p. \quad (19)$$

Replacing $A_p$ in eq. (16) with $L \cdot s \cdot C_0$, and eliminating $s$ with eq. (18), the effective diffusivity for scavenging effect is expressed by eq. (19). The magnitude of $L$ value is dependent of dislocation velocity, solute concentration in matrix, temperature and the size misfit of solutes.\(^{25,26}\) however, it remains constant as long as the dislocation velocity remains unchanged. Using $D_l$ for grip parts and $D'$ for gauge parts, the enhanced coarsening of precipitates in Al–Cu alloy,\(^{14}\) Mod.9Cr–1Mo steel\(^{22,23}\) and magnesium alloy of AZ80\(^{24}\) can be explained reasonably by the scavenging effect. The detail will be published elsewhere.
5. Summary

The effect of creep on the coarsening of precipitates was investigated of an extruded commercial magnesium alloy AZ80 at 433 K. The measurements of precipitate diameter revealed that the coarsening rate of precipitates was higher at gauge parts than at grip parts. The analysis on the experimental basis showed the enhanced growth of precipitates during creep was not due to creep stress but to creep strain. The results suggest that the particle growth is assisted by the scavenging effect of dislocations.

REFERENCES