A Creep Analysis of Superplastic Materials*
(Constant Load Creep of the Al-Zn Eutectoid-Base
Superplastic Alloys)

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Some phenomenological equations to hold among creep parameters of superplastic materials have been derived from a mechanical equation of state for superplasticity. A constant tensile-load creep test was carried out for Al-Zn eutectoid-base alloys to verify these equations. The equations agreed fairly well with the experimental results. Principal results obtained are as follows: Steady state creep disappears under the constant load. Accelerated creep accounts for most of creep strain. The relation between creep strain and time, and the time to rupture will be obtained if the minimum creep strain rate and m -value are known, where m is the strain rate sensitivity.

Key Words : Plasticity, Metallic Materials, Creep, Superplasticity, Constant-Load Test, Grain Size Dependence

1. Introduction

Recently the technology of forming materials making use of a dramatic ductility caused by superplasticity has attracted special interest. It is expected that forming and fabricating methods fully utilizing this phenomenon will certainly dominate the field of production technology in the near future⁽¹⁾. Various basic and applied studies have been carried out to realize this sort of fabricating technology and have accumulated a number of data gradually⁽²⁾. It is quite probable that the fabricated products in practical use will generally be used under some load and thereupon undergo a creep. However, most of studies reported on the creep behavior of superplastic materials have been metallographic ones⁽³⁾⁻⁽⁶⁾, and analyses which can be used for practical creep designs have been very few. Superplastic materials show strong strain rate dependence but do not strain-harden during the deformation⁽⁹⁾. Their creep behavior is therefore expected to be different from that of ordinary metallic materials.

In this study as the first stage to establish the theory of creep deformation for superplastic materials, phenomenological formulas for the creep will be derived from a mechanical equation of state for superplasticity. Then, in order to examine the formulas experimentally, constant-load creep tests by uniaxial tension will be

made of Al-Zn eutectoid-base superplastic alloys which have been used for various basic studies and put to commercial uses. Moreover, since the grain size dependence of deformation stress etc. of the superplastic material is different from that of ordinary metallic materials⁽¹⁰⁾ ⁽¹¹⁾, the effects of grain size and additional elements will also be examined.

2. Basic Equations for Creep Phenomenon of Superplastic Materials

Under superplastic conditions, the deformation (flow) stress σ is generally given by⁽¹²⁾

$$\sigma = K \dot{\epsilon}^m \quad \dots\dots\dots (1)$$

Alternatively, Eq.(1) can be expressed by

$$\dot{\epsilon} = K' (\sigma^n / L_0^a) \exp(-U/RT) \quad \dots\dots (2)$$

where $\dot{\epsilon}$ is the true strain rate; m , the strain rate sensitivity; L_0 , the initial grain size; U , the activation energy; R , the gas constant; T , the absolute temperature; $n=1/m$, K , K' and K'' are constants, and a is an exponent. In the case of constant-stress creep, since σ is a constant in Eq.(2), we obtain

$$\epsilon = K'' \sigma^n t \quad \dots\dots\dots (3)$$

This equation indicates a steady state creep, where ϵ is the true strain and t , the creep time. Equation (3) is verified empirically by the fact that the steady state creep appears over a considerably wide range of strains under constant stress in the superplastic materials, which has already been observed by Nishimura et al.⁽⁵⁾

In the case of constant load creep, σ is a function of nominal strain e , and

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hence of the time t . Therefore, $\sigma = F(1+e)/S_0$, where F is the load and S_0 , the initial cross sectional area. Since the relation $\epsilon = \ln(1+e)$ holds between the true and the nominal strains, we have

$$\sigma = \sigma_I \exp \epsilon \quad \dots\dots\dots(4)$$

where $\sigma_I (=F/S_0)$ is the initial creep stress. The strain distribution along the gauge length of a specimen is assumed to be uniform. From Eqs. (1) and (4), we obtain $\epsilon = A^n \exp(n\epsilon)$ where $A = \sigma_I/K$. From this we have

$$\int \exp(-n\epsilon) d\epsilon = A^n t + C' = (-1/n) \exp(-n\epsilon)$$

where $C' = -1/n$, since $\epsilon = 0$ at $t = 0$. Thus

$$\epsilon = (1/n) \ln(1 - n A^n t)^{-1} \quad \dots\dots(5)$$

From Eq.(5), we obtain

$$\dot{\epsilon} = A^n / (1 - n A^n t) \quad \dots\dots(6)$$

By eliminating the time t from Eqs. (5) and (6), ϵ is expressed as:

$$\epsilon = \ln(1/A) + m \ln \dot{\epsilon} \quad \dots\dots(7)$$

This Eq.(7) is exactly the same expression as that introduced by Nishimura et al.⁽⁸⁾. In an earlier stage of deformation, i.e., $t \ll 1$, by expanding the right-hand side of Eq. (5) in a series and neglecting terms of order two and more in the series, we obtain

$$\epsilon \approx A^n t \quad \dots\dots\dots(8)$$

This denotes the steady state creep. However, the steady state creep cannot exist in the case of superplastic deformation under the constant load. The reason is as follows: If we put $\epsilon = \epsilon_s = \text{const.}$ in Eq. (7), i.e., assume the condition as steady state creep, where $\dot{\epsilon}_s$ is the strain rate at the steady state, then the right hand side of Eq.(7) becomes constant, and consequently ϵ becomes constant, which means the deformation does not progress. Thus there arises a contradiction. Only in the initial stage of deformation, i.e., when $t \ll 1$, the relation $\epsilon = \epsilon_s t$ holds, which shows a similar behavior to the steady state creep.

Tensile tests has shown that a region of rapid work hardening appears in an earlier stage of deformation until a stationary superplastic flow begins⁽¹³⁾, which probably brings about the transient creep. This region is, however, so narrow compared with the total strain that the superplastic flow starts immediately⁽¹⁰⁾. The period of the transient creep is therefore very short and the deformation which satisfies Eq. (5) may start immediately. Nishimura et al.⁽⁹⁾ have also observed that the region of the transient creep is quite narrow. Consequently, the accelerated creep is expected to start immediately after the transient creep, which results in an inflection point appearing on its creep curve, and that the accelerated creep accounts for most of the deformation in which characteristics of superplasticity appear. In this report, discussions will be performed mostly on this accelerated creep, and the problems on the transient creep will be discussed at another opportunity.

Now, Eqs.(5) to (8) are significant

only in the region of superplastic flow, since they are derived from Eq.(1). The values of A^n in Eq.(8) correspond therefore to the strain rate at the onset point of a superplastic flow. The creep strain rate becomes minimum when $t=0$, since n is ≥ 0 and A^n is > 0 in Eq.(6). Thence, A^n is believed to be equal to the minimum creep rate $\dot{\epsilon}_{\min}$, since

$$(\dot{\epsilon})_{\text{minimum}} = A^n \equiv \dot{\epsilon}_{\min} \quad \dots\dots(9)$$

As $\dot{\epsilon}_{\min}$ corresponds with $(\partial \epsilon / \partial t)$ at which the ϵ vs t curve shows an inflection point, this moment may agree with the onset of a superplastic flow.

The time to rupture t_r can be obtained by summing up staying times at each strain-rate stage until the rupture occurs, that is, $t_r = \sum t_i$. If $\lim_{t \rightarrow \infty} t_i \neq 0$, then t_r diverges toward infinity. Since from Eq.(6)

$$t = 1/n A^n - 1/n \dot{\epsilon} \quad \dots\dots(10)$$

we have $(\partial t / \partial \dot{\epsilon}) = 1/n \dot{\epsilon}^2$. For convergence of t_r , the time t at which $(\partial t / \partial \dot{\epsilon})$ converges to zero must exist. Consequently, we have $n \dot{\epsilon}^2 \rightarrow \infty (n \dot{\epsilon} \rightarrow \infty)$. The rupture time t_r is given by $t_r = 1/n A^n$ from Eq.(10), since the rupture occurs when $n \dot{\epsilon} \rightarrow \infty$. As $A^n = \dot{\epsilon}_{\min}$, we obtain

$$t_r = 1 / n \dot{\epsilon}_{\min} = m / \dot{\epsilon}_{\min} \quad \dots\dots(11)$$

The rupture life is therefore calculated from Eq.(11), if m and $\dot{\epsilon}_{\min}$ values are known. The same formula as Eq.(11) is also obtained by assuming that the rupture occurs when the cross section of a specimen becomes zero and therefore $e \rightarrow \infty$, since $t = \{1 - 1/(1+e)^n\} / n A^n$ from Eq.(5). Let's consider which of the factors diverges to infinity, n or $\dot{\epsilon}$. From Fig.1, which will be shown later, we always have $\dot{\epsilon} \gg 1$ at rupture points on ϵ vs t curves but it does not diverge to infinity. Hatayama et al.⁽¹⁴⁾ have supposed that a finite strain rate which induces $m=0$ should exist even when the rupture occurs. From this idea, it appears that $n=1/m=1/0 \rightarrow \infty$ rather than $\dot{\epsilon} \rightarrow 0$. The effects of stress, temperature, and grain size on the minimum creep rate are, from Eqs. (1),(2) and (9), given by

$$\begin{aligned} \dot{\epsilon}_{\min} &= A^n = (\sigma_I / K)^n = \\ &= K' (\sigma_I^n / L_0^a) \exp(-U/RT) \quad \dots\dots(12) \end{aligned}$$

Though grains are thought to grow during superplastic deformation⁽¹⁵⁾, only the initial grain diameter L_0 is taken into consideration here. The effect of grain growth will be discussed later. From Eq.(12) the following expressions are obtained: under the conditions that both L_0 and T are constants,

$$\ln \dot{\epsilon}_{\min} \propto n \ln \sigma_I \quad \dots\dots(12-a)$$

When σ_I and L_0 are constants,

$$\ln \dot{\epsilon}_{\min} \propto -1/T, \quad \dots\dots(12-b)$$

When T and σ_I are constants,

$$\ln \dot{\epsilon}_{\min} \propto a \ln(1/L_0) \quad \dots\dots(12-c)$$

On the other hand, the effects of σ_I , T , and L_0 on the time to rupture are, from Eqs.

(1), (2) and (11), given by

$$t_r = K'^{-1} m L_0^a \sigma_I^{-n} \exp(U/RT) \dots (13)$$

In the same manner as described above, the following expressions may hold :

$$\ln t_r \propto -n \ln \sigma_I, L_0, T = \text{const.} \quad (13-a)$$

$$\ln t_r \propto 1/T, \sigma_I, L_0 = \text{const.} \quad (13-b)$$

$$\ln t_r \propto a \ln L_0, T, \sigma_I = \text{const.} \quad (13-c)$$

Now let us derive an expression between the initial stress and the time as a function of creep strain to make a creep design diagram. From Eqs. (5) and (2) or (12) we have $t = K'^{-1} m L_0^a \sigma_I^{-n} \exp(U/RT) \{1 - 1/(1+e)^n\}$ and therefore

$$\ln \sigma_I = -m \ln t + m \ln [K'^{-1} m L_0^a \exp(U/RT) \times \{1 - 1/(1+e)^n\}] \dots (14)$$

This gives a basic expression for the creep design diagram of the superplastic materials. As it is difficult to detect the effect of temperature from the creep design data diagram, we will introduce the Larson-Miller parameter to obtain a master rupture curve. From Eq. (13), under constant initial stress, we have $t_r^{-1} = B \exp(-U/RT)$. Therefore,

$$T (C + \log t_r) = U / 2.3 R \dots (15)$$

where, $C = \ln B = \ln [K'^{-1} m L_0^a \sigma_I^{-n}] - 1$.

As described above, we have derived correlative equations among various deformation parameters expected in the creep of superplastic materials on the basis of Eqs. (1) and (2). In Chapter 4 these equations will be compared with experiment in order to verify their validity and usefulness.

3. Materials Used and Experimental Procedures

Table 1 Chemical compositions of specimens (in weight %)

	Al	Cu	Mg	Fe	Zn
SPZ	21.22	0.01	<0.001	0.011	bal.
SPZ 1	22.09	0.13	0.005	0.016	bal.

Materials used are eutectoid-base Al-Zn binary (SPZ) and Al-Zn-Cu ternary (SPZ1) alloys and their chemical compositions are shown in Table 1. The alloys were melted in air, and cast into a graphite mold. After homogenization at 648K, the alloys were quenched into iced water, then hot rolled and finally warm rolled to 1 mm thick plate. Specimens, gauge length 40 mm, gauge width 5 mm, were made by mechanical machining from the plate with the tensile axis parallel to the rolling direction. Specimens were then subjected to solution treatment at 648K for 3h and water quenching to make their structure equi-axed. Specimen having various initial grain sizes were obtained by aging at 473K for a period ranging from 0.25 to 480h. Each specimen was kept at 243K until just before the test was performed. Average grain sizes were determined from scanning electron micrographs by measuring the mean linear intercept $\bar{L}$, which was converted to the average spatial grain size L , using the relationship $L = 1.75 \bar{L}$ (16).

Creep tests were carried out on a direct load type testing machine at initial stresses from 9.81 to 196MPa and at temperatures from 288 to 473K. This temperature range corresponds to 0.4 to 0.7 T_m , where T_m is the melting point. The temperature of the specimens during deformation was measured at both upper and bottom sides of them. A holding time of the specimen before testing was set such that grains hardly grow during the time. The temperature difference in longitudinal direction of the specimen was within $\pm 2K$ and its fluctuation during the deformation was within ± 0.5 per cent of the test temperature. Electric displacement meter was used to measure the elongation of the specimen. Microstructures before and after the deformations were observed by making use of scanning electron microscopy.

4. Experimental Results and Discussions

Creep curves clearly show a feature of superplasticity as shown in Fig. 1, from which it is seen that an accelerated creep region is markedly wide, while the transient creep is quite narrow. The relation ϵ

Fig. 1 True strain vs creep time curves for SPZ under various constant loads

vs t was replotted on a log-log scale to magnify the initial creep region well as shown in Fig. 2. It is confirmed from the figure that there is a quasi-secondary creep in the initial stage as indicated by Eq. (8) since $\epsilon \propto t$ until $\epsilon \approx 0.02$. Now, the effects of σ_I , T , and L_0 on the minimum creep rate are examined to verify validities of Eqs. (12.a) to (12.c). From Fig. 3

showing the relationship $\ln \dot{\epsilon}_{\min}$ vs $\ln \sigma_I$, it is seen that the slope of the curve $n' (= \partial \ln \dot{\epsilon}_{\min} / \partial \ln \sigma_I)$ decreases with a fall in strain rate, indicating that the n -value depends on strain rate⁽¹⁴⁾. From this fact, we may conclude that Eq.(12.a) agrees with the experiment. It is seen from Fig.4 showing the effects of σ_I , L_0 and T on the n -

value and consequently on the m -value that the m -value has a tendency to increase with decreasing σ_I and hence decreasing $\dot{\epsilon}$, and increasing T . This tendency is in good agreement with the result of tensile test⁽¹⁷⁾. Figure 5 showing the relationship $\ln \dot{\epsilon}_{\min}$ vs $1/T$ indicates that the slope of the curves alters at temperatures around

Fig. 2 True strain vs creep time on a log-log scale

Fig. 3 Effect of initial stress on the minimum creep rate for SPZ (a) and SPZ1 (b)

320K. Since the activation energy U of the deformation process is given by multiplying the slope by the gas constant R , this fact implies that the value of U alters at around 320K. This is the same tendency observed in the tensile test⁽¹¹⁾. Since the slopes of the lines decrease with increasing initial stress, the activation energy obtained must be the observed value. As the relationship $U = U_0 - V f(\sigma)$ holds between U and the true activation energy U_t ⁽¹⁸⁾, an increase in stress results

in a decrease in the U -value, where V is the activation volume and $f(\sigma)$ is a function of stress. The tendency described above may therefore be explained by this idea. Equation (12.b) is then believed to agree with the experiment, though qualitatively. Grain size dependence of $\dot{\epsilon}_{\min}$ was found to be such that $\dot{\epsilon}_{\min}$ decreases with an increase in grain size at each temperature as shown in Fig.6 (a) and (c). This fact suggests that grain growth will raise creep resistance. The gradients of the lines vary from approximately 1.0 to 2.0 depending on σ_I and T . Though the data show a scattering to some extent, Eq.(12.c) may be satisfied well.

The dependences of σ_I , T and L_0 on the time to rupture t_r are respectively shown in Fig.7, Fig.8, and Fig.6(b) and (d). The values of t_r has a tendency to increase with a decreases in stress and in temperature. More-

over, t_r increases with growing initial grain size. This clearly shows that the grain growth is fairly effective for raising the rupture life of the alloy from the view point of its internal structure. These experimental results (Figures 7, 8 and 6 (b)(d)) agree well with the expressions (13.a) to (13.c). Thus, it is believed that Eqs.(12) and (13) agree well with and are

Fig. 4 Effect of initial stress on m-value

Fig. 5 The minimum creep rate as a function of the reciprocal of absolute temperature

verified by the experiment, though semi quantitatively.

The time dependence of the creep strain is established from Eq.(5) if n - or m -value and $\dot{\epsilon}_{min}$ ($=A\sigma^n$) are known. Hence, we can present a theoretical curve which is shown in Fig.2 as a solid lines, by sub-

Fig. 6 Effects of grain size on the minimum creep rate and the time to rupture for SPZ (a) and (b), and SPZ1 (c) and (d), respectively

Fig. 7 Effects of initial creep stress on the time to rupture for SPZ (a) and SPZ1 (b), respectively

Fig. 8 Rupture time as a function of the reciprocal of absolute temperature for SPZ

stituting values of n and $\epsilon_{\min}$ obtained in this experiment into Eq.(5). Calculated values appear to be in good agreement with those obtained experimentally when stress is high and consequently when the rupture time is short. However, as σ_i decreases, their agreement lowers and the observed value of t_r increases towards longer side of the time than the calculated value. This may be due to the effect of grain growth which will occur during the creep deformation in which a very long time will be taken up to a rupture. Then the grain growth during creep deformation was observed. It was found that with increasing strain the grain size grows and it becomes twice as large as the original size after 130 per cent elongation as shown in Fig.9. The grain growth brings about a decrease in $\epsilon_{\min}$ and an increase in t_r , as was already shown in Fig.6. Therefore, let us consider in detail the effect of grain-growth during

the deformation. From Eq.(12) and the conversion of Eq.(5) into a new expression containing the grain size dependence explicitly, we obtain

$$\epsilon = (1/n) \times \ln(1 - PL_0^{-a} t)^{-1} \quad \dots (16)$$

where $P = K'n\sigma_i^2 \exp(-U/RT)$. As the grain size become a variable in this case, L_0 should be replaced by L which has been proposed as $L = L_0 + kt^c$ during the superplastic deformation (19). The relationship between the creep strain and the time as a function of grain growth is therefore given by

Fig. 9 Characteristics of microstructures after 130 per cent creep elongation deformed at 288K; (b). Stress was applied in the vertical direction. An initial micro-structure is also shown; (a), $L_0 = 0.67 \mu\text{m}$

$$\epsilon = (1/n) \ln \left[1 - P \int_0^t (L_0 + k t^c)^{-a} dt \right]^{-1} \quad \dots (17)$$

The rupture time based on Eq.(5) or Eq.(16) is, from Eq.(11), $t_r = m/\dot{\epsilon}_{\min} = L_0^a/P$, whereas t_r based on Eq.(17) is given by

$$t_r' = (L_0 + k t^c)^a / P \quad \dots (18)$$

Fig. 10 Variation of $t_r \times \dot{\epsilon}_{\min}$ with initial stress

Fig. 11 Effect of initial stress on total elongation

Since $c > 0$, $k > 0$, and $a > 0$ concerning the superplastic alloy as mentioned above, it is seen from Eq.(18) that the grain growth becomes greater with an increase in the time of creep, and consequently the time to rupture also increases. The rate of prolongation of the time to rupture induced by the grain growth is therefore given by $t_r'/t_r = (1 + k t^c/L_0)^a$. Quantitative evaluation of Eq.(18) will not be treated in the present paper since the values of a , c , and k are still uncertain. It can be concluded from these results that the increase of the actual rupture time from the value expected is attributable to the effect of dynamic grain growth during the deformation. In the case of a short lifetime creep, since the grain growth is not sufficient and the value of L can be regarded as approximately constant during the deformation, Eq.(5) would be able to explain comparatively well the experimental results.

Since $m = t_r \times \dot{\epsilon}_{\min}$ from Eq.(11), the m -value will be obtained if $\dot{\epsilon}_{\min}$ and t_r are known. Thereupon the value of $t_r \times \dot{\epsilon}_{\min}$ was evaluated and shown in Fig.10. It is seen from comparing Fig.10 with Fig.4 that, in a high-stressed region at 288K, the agreement between m and $\dot{\epsilon}_{\min} \times t_r$ values is comparatively good since $m \approx t_r \times \dot{\epsilon}_{\min}$, whereas always $t_r \times \dot{\epsilon}_{\min} > m$ in a low-stressed region at elevated temperatures. This may be due to the grain growth during the deformation which is, as stated above, contributable to the increase of its life. It appears that a reason why the formula $t_r = m/\dot{\epsilon}_{\min}$ is a good approximation for an estimation of the rupture time is such that the accelerated creep where the alloy flows superplastically accounts for most of total

Fig. 12 Effect of initial grain size on total elongation

creep strain and therefore the contribution of the transient creep to the total strain is negligibly small.

As a striking feature of superplasticity is its remarkable elongation, the effects of σ_T , T , and L_0 on the elongation, e_t were examined as shown in Fig.11. It is seen from the figure that e_t increases markedly with decreasing σ_T . This tendency is reverse to that of ordinary metallic materials. The reason may be that, since the m -value increases with decreasing σ_T (or decreasing strain rate) as already shown in Fig.4, necking resistance of a specimen rises at lower stresses and consequently a large elongation results. Moreover, the value of e_t increases consistently with temperature, but hardly depends on grain size as shown in Fig.12. It can be concluded that a coarsening of initial grain size reduces $\dot{\epsilon}_{min}$ without a decrease in elongation and increases the rupture time markedly, and hence it is practically

very effective to improve the creep strength. Creep rupture strength was examined for a creep design of the superplastic alloys. Figure 13 shows a design data diagram obtained. Initial stress σ_T relates linearly with t on a log-log scale, which indicates that Eq.(14) may, though qualitatively, be valid as shown in Fig.13(a). The lines show discontinuities at a time though the moment alters depending on temperature and grain size. This fact suggests that the deformation process or rupture mechanism changes around the moment. Nishimura et al. (8) have applied a linear approximation to the σ vs $\log t$ relation. Then we replotted the present results on a semi log scale, but the linear approximation was seen to be insufficient as shown in Fig.13(b). The shape of the curves resembles that of an ordinary material such as superalloy S-590 (20), and distinctive features of superplasticity are hardly seen. The SPZ1 also shows a similar tendency.

The C-value in the Larson-Miller parameter was estimated as 6.0 from Fig.8, and has been applied to make a master rupture curve. The value of 6.0 is approximately one third of that of steels, indicating a feature that the C-value of the superplastic alloys is fairly small. The master rupture curve moves to the upper side of the diagram with an increase in initial grain size as shown in Fig.14, which implies that the grain growth raises its creep strength. It is well known that these curves are convenient for creep design, since there exists equivalence between temperature and the rupture time at a constant stress.

Now, Fig.15 was plotted to examine how much a coarsening of initial grain size can improve the values of t_r and $\dot{\epsilon}_{min}$. The initial values of L_0 are 0.67 μm and 0.60 μm for SPZ and SPZ1, respectively, and the initial values of t_{r0} and $\dot{\epsilon}_{min0}$ are those corresponding to these grain sizes, respectively. It is evi-

Fig. 13 Design data diagrams on a log scale (a) and a semilog scale (b), respectively

Fig. 14 Master rupture curve for SPZ

dent that growing the initial grain size, for example, four times as much as the L_0 -value does improve the values of t_r and $\dot{\epsilon}_{min}$ to about 4 to 7 times and 0.2 to 0.7 times their initial values, respectively. From comparison of t_r and $\dot{\epsilon}_{min}$ in SPZ with those in SPZ1 in the case they have the same initial grain size ($L_0 = 1 \mu m$) at room temperature, it is seen that the creep strength is markedly improved by the addition of 0.13 per cent Cu since t_r and $\dot{\epsilon}_{min}$ in SPZ1 change to about 20 and 0.045 times when $\sigma_1 = 49 \text{ MPa}$, and 5.5 and 0.2 times when $\sigma_1 = 172 \text{ MPa}$ their initial values, respectively. Furthermore, in order to examine a combination effect of the grain growth and the additional elements, t_r and $\dot{\epsilon}_{min}$ in the case of $L_0 = 0.67 \mu m$ for SPZ are compared with those in the case of $L_0 = 2.5 \mu m$ for SPZ1 at $\sigma_1 = 49 \text{ MPa}$, and it is seen that t_r assumes about 60 times and $\dot{\epsilon}_{min}$ does 0.013 times the original values, which means that the creep strength has been greatly improved. The additional elements, however, act to lower superplasticity. Therefore, it can be recommended that the additional elements be controlled practically to their amount within which they would not lower the superplasticity and that an appropriate heat-treatment to grow the initial grain size should be carried out at reasonable cost. Moreover, it should be noted that the grain growth during creep deformation brings about the improvement of the creep strength as the result. At any rate, in creep under the conditions where superplasticity arises, the onset of an accelerated creep comes at earlier stage of deformation, in which the strain increases quite acceleratedly with an increase in stress. Hence, such superplastic materials may be inappropriate as structural materials to be used for a long period. It may therefore be necessary to use the material under such conditions where no superplasticity will appear, e.g. use at lower temperatures, or to destroy its superplastic structure by means of heat treatment for instance by changing it into a lamellar structure after superplastic forming. It should be pointed out finally that the characteristic of the creep of superplastic materials consists in that its accelerated

Fig. 15 Variations of the ratios of t_r and $\dot{\epsilon}_{min}$ to their initial values with coarsening of grain size

creep region is very wide though its transient creep region is not always narrow as compared with the case of ordinary materials. Dynamic equilibrium exists also in this accelerated creep region⁽⁸⁾. Therefore, the region of the accelerated creep of superplastic materials cannot be regarded as the same one as the tertiary creep of ordinary metallic materials. It will be much better to divide it into the region of the stationary accelerated creep and that of the rapid accelerated creep where plastic instability appears after the breakdown of dynamic equilibrium.

5. Conclusions

Creep deformation was analyzed phenomenologically on the basis of the mechanical equation of state for superplasticity, and formulas which would hold among creep parameters were derived as expressed by Eqs. (3) to (18). In order to examine these equations experimentally, tensile creep tests at constant load were applied to Al-Zn eutectoid base alloys and it was found that these equations were in good agreement with experimental results, if we took the grain growth during the deformation into account. The principal results obtained are as follows:

(1) The steady state creep is lost under constant load, which induces the appearance of an inflection point on creep-rupture curve, since transient creep is immediately followed by accelerated creep. Moreover, the accelerated creep, where superplastic features may markedly appear, accounts for most of creep strain, whereas the period of the transient creep is actually short.

(2) The relation between creep strain and time can be obtained if m -value and $\dot{\epsilon}_{min}$ are known. Similarly, the time to rupture can be given by the expression $t_r = m / \dot{\epsilon}_{min}$.

(3) Coarsening of initial grain size

is quite effective to improve creep strength of the alloy, since it markedly increases t_r and at the same time decreases $\dot{\epsilon}_{min}$. Furthermore, an addition of 0.13 per cent Cu extremely raises the creep strength. The grain coarsening and the addition of copper have a multiple effect for improving the creep strength.

(4) Dynamic grain growth promoted by stresses during deformation results in a long life.

(5) Larson-Miller parameter C is approximately one third that of ordinary metals such as steel.

(6) Superplastic materials may not be fit as structural materials which will be used for a long period.

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