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Interpretation of superplastic flow in terms of a threshold stress

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In several recent experiments on the Zn–22% Al eutectoid and the Pb–62% Sn eutectic, a sigmoidal relationship between stress and strain rate is noted and the mechanical behaviour has been divided into three regions: low-stress region (region I), intermediate-stress region (the superplastic region or region II), and high-stress region (region III). In region II, the stress exponent, n, is $\simeq 2$ and the apparent activation energy, Q, is close to grain-boundary diffusion, Q_{gb}, but in both regions I and III the stress exponent and the activation energy increase ($n > 2$ and $Q > Q_{gb}$). Analysis of the experimental data of the two superplastic alloys suggests that the transition in behaviour between region II and region I may not necessarily reflect a change in deformation process but can arise from the presence of a threshold stress which decreases strongly with increasing temperature. Based on consideration of various possible threshold stress processes during superplastic flow, it seems most likely that a threshold stress which depends strongly on temperature may result from impurity atom segregation at boundaries and their interaction with boundary dislocations.

1. Introduction

Micrograin superplasticity, which refers to the ability of ultra fine-grained metals and alloys to exhibit upon deformation at elevated temperatures ($> 0.4T_m$, where T_m is the absolute melting point of the material) extremely large, neck-free elongations of several hundreds of per cent, has been extensively investigated in recent years. As a result of these investigations, it has been established that superplasticity is a diffusion-controlled process which can be well represented by the normalized form of the Dorn equation given by [1]

$$\frac{\dot{\gamma}kT}{DGb} = A\left(\frac{b}{d}\right)^s \left(\frac{\tau}{G}\right)^n \quad (1)$$

where $\dot{\gamma}$ is the steady-state strain rate, k is Boltzmann's constant, T is the absolute temperature, D is the appropriate diffusion coefficient, G is the shear modulus, b is the Burgers vector, A is a dimensionless constant, d is the grain size, s is the grain size sensitivity, τ is the applied stress, and n is the stress exponent. Additionally, several experimental investigations [2–4], which were conducted over ranges of stress, temperature and grain size, have revealed the presence of a sigmoidal relationship between the steady-state strain rate, $\dot{\gamma}$, and the applied stress, τ in two superplastic alloys: the Zn–22% Al eutectoid [2, 3] and the Pb–62% Sn eutectic [4]. This sigmoidal relationship is manifested by the presence of three distinct deformation regions; at very low stresses (region I) the stress exponent, n, is 3 to 4.5; at intermediate stresses (region II or the superplastic region) the stress exponent, n, is close to 2; and at high stresses (region III) the stress exponent, n, again increases.

It was argued [5, 6] that the high stress exponent in region I may represent a false value arising from the occurrence of concurrent grain growth during the test. However, several pieces of experimental evidence [7–10] tend to refute that argument. These include the observation of a well-defined region I with a high stress exponent in creep experiments involving negligible grain growth [8, 9], the insensitivity of the value of the stress exponent in region I to the order in which the experimental data were obtained [7] – whether the specimen was cycled during testing by increas-

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ing or by decreasing the applied stress – and the close similarity in neck behaviour [10] between region I and region III which also exhibits a high stress exponent.

The increase in the value of the stress exponent of Zn–22% Al and Pb–62% Sn at low stresses represents a genuine trend and two possible explanations for this increase were offered [2–4, 11–15]. First, it was suggested [2–4] that region I may be associated with the emergence of a new deformation mechanism which interacts sequentially with the mechanism controlling region II. This suggestion is based on two observations: (a) the experimental data of Zn–22% Al [2, 3] and Pb–62% Sn [4] in region I, when plotted in the normalized form of the Dorn equation [1], tend to fall very close to single straight lines [16]; and (b) the increase in the value of the stress exponent for both alloys from $n \simeq 2$ (region II) to $n \simeq 3.5$ (region I) with decreasing stress is accompanied by a corresponding increase in the value of the activation energy for superplastic flow [2–4]. Observation (a) is generally viewed as evidence of the dominance of a single deformation process and observation (b) appears, according to the results of a recent analysis [17], consistent with the presence of a sequential interaction between two deformation mechanisms. While the interpretation of region I in terms of a concurrent sequential process seems attractive, attempts [7, 18, 19] have so far been unsuccessful to develop a deformation mechanism that can account for all of the mechanical characteristics noted in that region (high stress exponent, high activation energy and strong grain-size dependence).

Second, it was pointed out [11–15] that region I may not be a consequence of a separate deformation process but may arise from the existence of a threshold stress, τ_0. This possibility, when incorporated into the development of two recent theories of superplasticity [14, 15], explains qualitatively the experimental sigmoidal behaviour of Zn–22% Al and Pb–62% Sn but fails to account for the higher activation energy observed in region I. Despite this failure, the interpretation of the low-stress superplastic flow in terms of a threshold stress needs to be re-examined in view of recent experimental and numerical results [20–22].

In the development of theories of superplasticity that introduce a threshold stress into the rate equation, it was postulated that the threshold stress is caused by processes that are insensitive to temperature [14, 15]. However, recent experimental measurements made during superplastic flow of a duplex stainless steel [20] revealed the presence of a threshold stress that decreased strongly with increasing temperature. By incorporating this experimental observation into hypothetical threshold-stress processes, it was demonstrated recently [21] that the operation of such processes would lead to an increase in both the stress exponent and the activation energy for deformation with decreasing stress, a trend which is essentially similar to those reported for Zn–22% Al and Pb–62% Sn at low stresses (region II and region I). More importantly, consideration of the nature of the interaction between different deformation processes suggested [21] that under certain experimental conditions $\log \dot{\gamma} - \log \tau$ plots (from which the stress exponent is inferred) and Arrhenius plots of $\log \dot{\gamma}$ against $1/T$ (from which Q is inferred) may not provide sufficiently good criteria to distinguish the difference between sequential processes and threshold stress processes that depend strongly on temperature. Additionally, a very recent analysis [22] of the effect of the operation of a threshold process on plastic instability showed that such a process, like two sequential processes, would result in a decrease in ductility with decreasing applied load. As documented elsewhere [23], the transition in the superplastic behaviour of Zn–22% Al at low stresses is also manifested by a decrease in ductility.

The above findings and results suggest that the concept of a threshold stress during superplastic flow is capable, at least qualitatively, of explaining the transition in the mechanical characteristics of superplastic materials from those of region II to those of region I. This suggestion motivated the work reported here, in which the experimental data of Zn–22% Al [2, 3] and Pb–62% Sn [4] are analysed to determine whether they can be quantitatively correlated with a threshold stress process and, if such a correlation exists, to speculate on the nature of the process.

2. Analysis

In an earlier paper [16], the experimental studies of Zn–22% Al and Pb–62% Sn in the superplastic region II were analysed in terms of Equation 1, and it was shown that all of the experimental data, when replotted in the normalized form of

$$\left(\frac{\dot{\gamma}kT}{D_{gb}Gb}\right)\left(\frac{d}{b}\right)^{s} \quad \text{against} \quad \frac{\tau}{G}$$

(for Zn–22% Al, $s = 2.4$ and for Pb–62% Sn, $s = 2.3$), can be well represented by

$$\frac{\dot{\gamma}kT}{D_{gb}Gb}\left(\frac{d}{b}\right)^{2.4} = 2.25 \times 10^{6}\left(\frac{\tau}{G}\right)^{2.3}, \quad (2)$$

and

$$\frac{\dot{\gamma}kT}{D_{gb}Gb}\left(\frac{d}{b}\right)^{2.3} = 1.3 \times 10^{5}\left(\frac{\tau}{G}\right)^{1.77}, \quad (3)$$

for Zn–22% Al and Pb–62% Sn, respectively. In Equations 2 and 3, D_{gb} represents the coefficient of grain-boundary diffusion.

As a first step in the analysis, it was assumed that the mechanical behaviour of both alloys, Zn–22% Al and Pb–62% Sn, in regions II and I can be described by Equations 2 and 3, respectively, with the exception that the applied stress, τ, is replaced by an effective stress τ_e; $\tau_e = \tau - \tau_0$, where τ_0 is the stress which opposes deformation. This assumption implies that at high stresses typical of those used in region II, the value of τ_0 is so small that the mechanical behaviour of the two materials can be reasonably approximated by Equations 2 and 3. However, the assumption does not specify the nature and origin of τ_0.

The second step involved an estimation of τ_0 using the experimental data of the two alloys. In estimating τ_0, two procedures were adopted. In the first procedure, the experimental results of Zn–22% Al and Pb–62% Sn obtained at different temperatures and a single grain size were plotted logarithmically as

$$\frac{\dot{\gamma}kT}{D_{gb}Gb} \quad \text{against} \quad \frac{\tau}{G}$$

in the same figure; D_{gb} and G were taken from [3, 4, 16]. As a result of such a plot, the data of each alloy in region II fall very close to a straight line, with a slope equal to the stress exponent (for Zn–22% Al $n = 2.3$ and for PB–62% Sn $n = 1.77$), but the data of region I deviate appreciably from the linear extrapolation of region II; they are located at stresses higher than those determined from the extrapolation of the line representing region II. The difference in stress level between that of a normalized datum point belonging to region I and that determined from the linear extrapolation of region II was considered, according to the assumption introduced in the first step, to represent τ_0. By applying the procedure to all experimental points obtained at the same temperature, an average value of τ_0 representing that temperature was estimated. The consistency of the procedure is demonstrated by the fact that the difference between the maximum and minimum values of τ_0 for a particular temperature is very small in most cases.

In the second procedure, the experimental data obtained for both alloys at a single temperature in regions II and I were replotted as $\dot{\gamma}^{1/n}$ (for Zn–22% Al $n = 2.3$ and for Pb–62% Sn $n = 1.77$) against τ on a double linear scale. The data points were then fitted with a single straight line whose extrapolation to zero strain rate gave the value of τ_0. Typical examples for application of this procedure to Zn–22% Al and Pb–62% Sn are given in Figs. 1 and 2, respectively. These examples clearly demonstrate that the data points of each alloy obtained at a single temperature in both regions II and I can be fitted nicely with a straight line.

Table I summarizes some of the results obtained from the above procedures. An examination of this table shows that the values of τ_0 estimated from both procedures not only are essentially identical but also are very sensitive to temperature; τ_0 decreases with increasing temperature.

A close inspection of the variation of τ_0 with temperature indicates that this variation is much stronger than that attributable to the modulus of elasticity. To develop an empirical relationship between τ_0 and T, the results for constant grain size were plotted using several trial functions. Based on an examination of these plots, the conclusion was that the variation of τ_0 with temperature can be best described by the following empirical equation:

$$\frac{\tau_0}{G} = B_0 \exp\frac{Q_0}{RT}, \quad (4)$$

where R ($= 8.31\,\mathrm{J\,mol^{-1}\,K^{-1}}$) is the gas constant, B_0 is a constant and Q_0 is an energy term. Two examples are given in Figs. 3 and 4 for Zn–22% Al and Pb–62% Sn, respectively, to illustrate the validity of Equation 4. In these two figures, $\log \tau_0/G$ is plotted against $1/T$ at constant grain size; for Zn–22% Al, $d = 2.3\,\mu\mathrm{m}$ and for Pb–62% Sn, $d = 5.8\,\mu\mathrm{m}$. By estimating Q_0 ($\mathrm{J\,mol^{-1}}$) and B_0 from the data of Figs. 3 and 4, Equation 4 can be expressed as

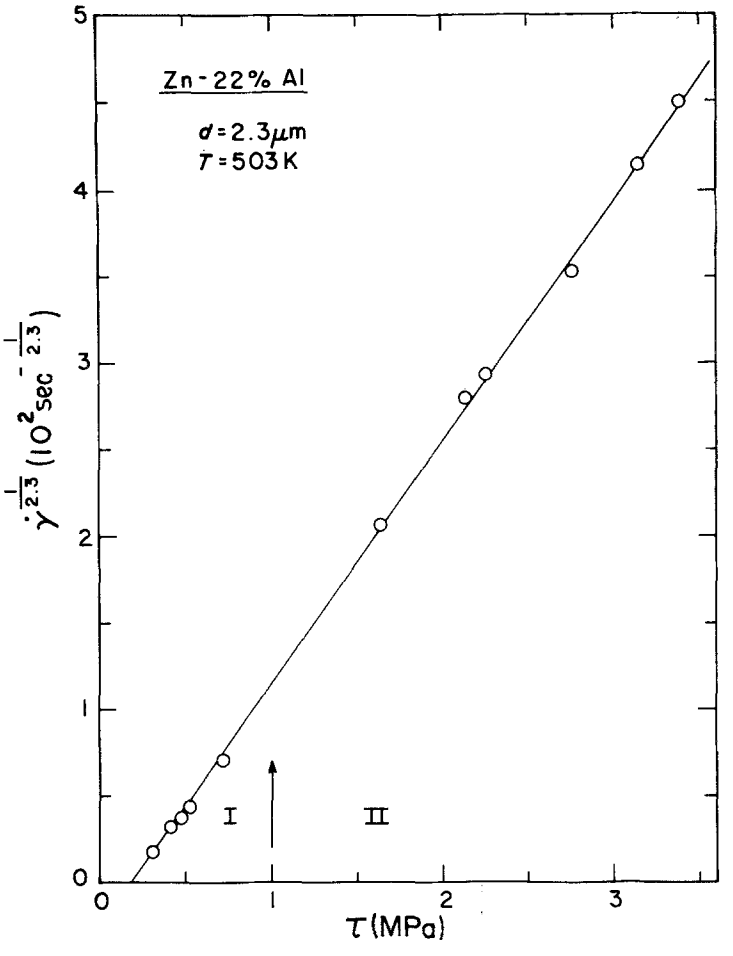

Figure 1 A plot of $\dot{\gamma}^{1/2.3}$ against τ for Zn–22% Al. Data were taken from [2] and [3].

$$\frac{\tau_0}{G} = 1.75 \times 10^{-8} \exp\left(\frac{2.38 \times 10^4}{RT}\right) \quad (5)$$

and

$$\frac{\tau_0}{G} = 5.75 \times 10^{-9} \exp\left(\frac{1.65 \times 10^4}{RT}\right), \quad (6)$$

for Zn–22% Al and Pb–62% Sn, respectively.

Having determined the functional dependence of τ_0 on T, attention was then focused on examining whether Q_0 and B_0 are sensitive to grain size. Consideration of the data for both alloys at other grain sizes showed that Q does not depend on d but B_0 (or τ_0/G) increases very slightly with increasing d; for example, an increase in d for Zn–22% Al by a factor of 3 is accompanied by

TABLE I Values of τ_0 for Zn–22% Al and Pb–62% Sn

Alloy	d (μm)	T (K)	τ_0 (10^2 MPa) from normalized data* Procedure 1	τ_0 (10^2 MPa) from $\dot{\gamma}^{1/n}$ against τ Procedure 2
Zn–22% Al	2.3	503	17	18
		463	30	34
		443	50	53
		409	70	70
Pb–62% Sn	5.8	422	1	1
		392	1.5	1.6
		361	2.3	2.4
		336	4.2	3.2

*τ_0 represents an average of several values.

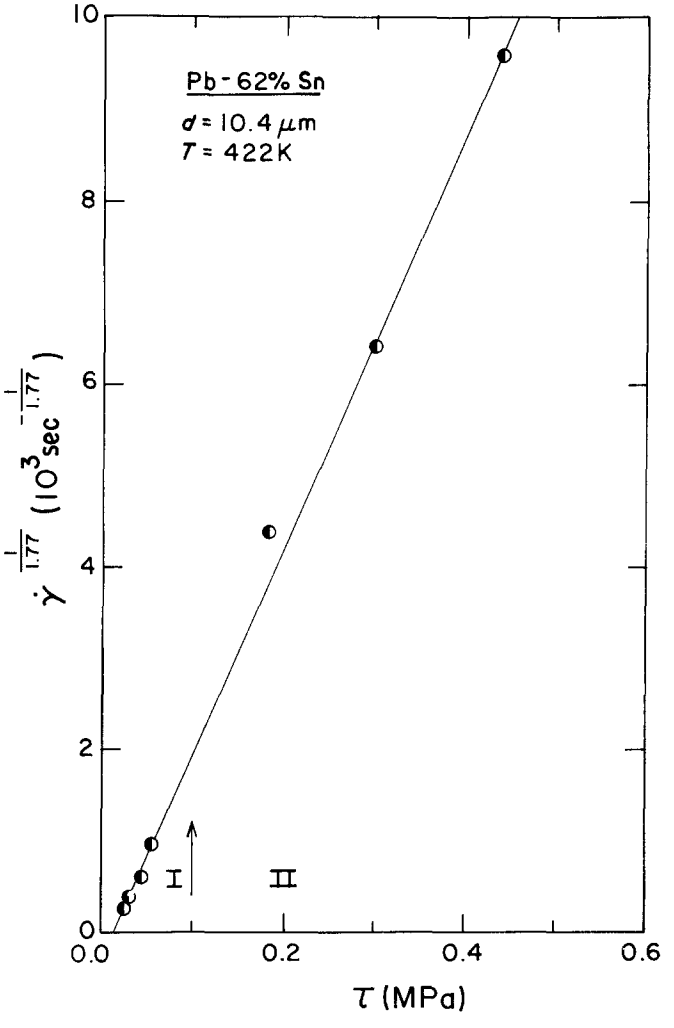

Figure 2 A plot of $\dot{\gamma}^{1/1.77}$ against τ for Pb–62% Sn. Data were taken from [4].

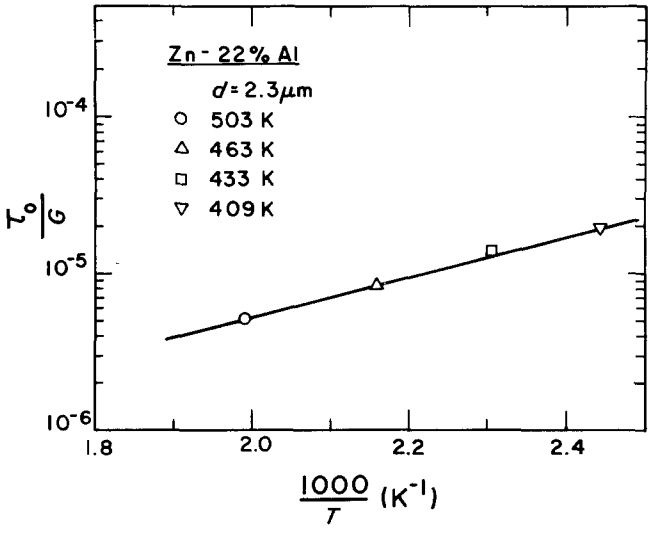

Figure 3 A plot of the log of the normalized threshold stress against $1/T$ for Zn–22% Al.

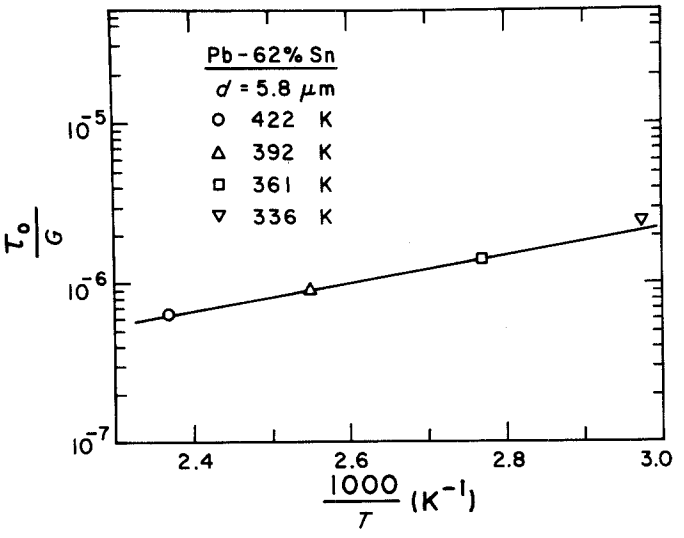

Figure 4 Variation of the log of the normalized threshold stress with $1/T$ for Pb–62% Sn.

15% increase in τ_0. In view of this very weak dependence of τ_0 on d, B_0 is taken as a constant whose value for Zn–22% Al and Pb–62% Sn is given by Equations 5 and 6, respectively.

Two observations are noted in regard to the expression of τ_0. First, the expression shows that the values of B_0 and Q_0 for Zn–22% Al are higher than those of B_0 and Q_0 for Pb–62% Sn. Second, Burton [13] also reported a tensile threshold stress of 0.18 MPa in his investigation on Pb–62% Sn at room temperature and it is tempting to check if that value can be predicted from Equation 6. By substituting $T = 300$ K and $G = 1.7 \times 10^4$ MPa in Equation 6, the estimated value of τ_0 is 0.075 MPa which, when multiplied by a factor of 2 to convert from shear to tension, compares favourably with that measured by Burton [13].

The validity of Equations 5 and 6 to the analysis of the experimental results of Zn–22%

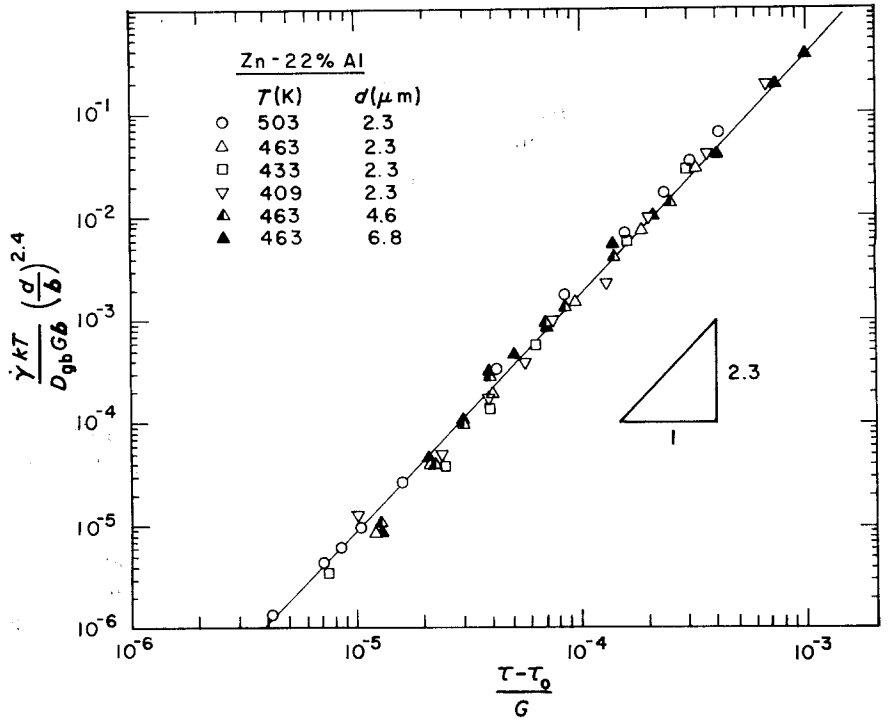

Figure 5 A plot of the log of the normalized shear strain rate against the log of the normalized effective stress ($= \tau - \tau_0/G$) for Zn–22% Al. Data were taken from [2] and [3].

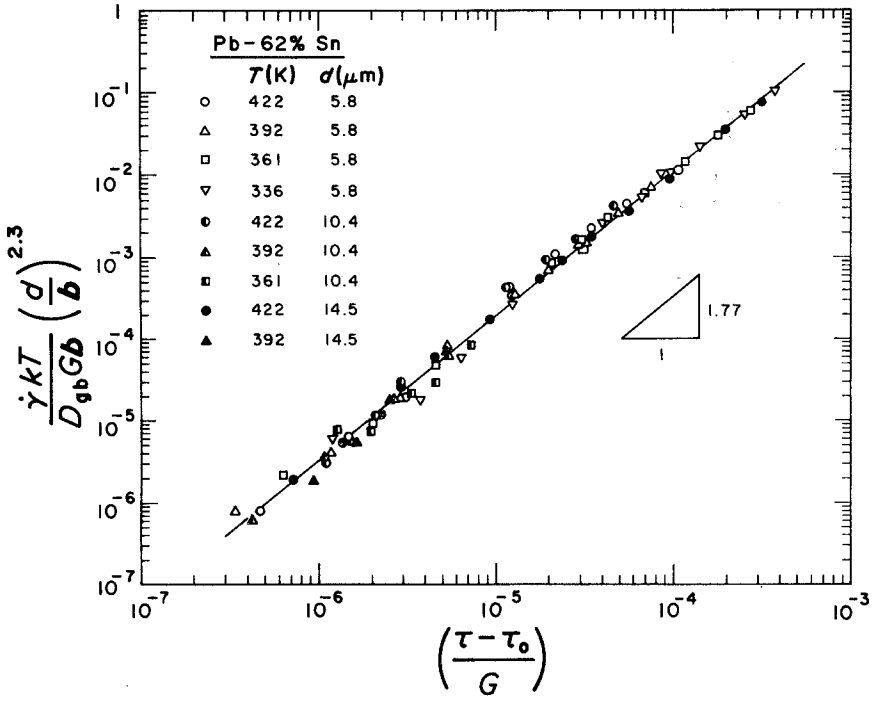

Figure 6 Dependence of log normalized shear strain rate on log normalized effective stress ($= \tau - \tau_0/G$) for Pb–62% Sn. Data were taken from [4].

Al and Pb–62% Sn in regions II and I was further examined by plotting all of the experimental data obtained for each alloy as

$$\frac{\dot{\gamma}kT}{D_{gb}Gb}\left(\frac{d}{b}\right)^{s} \text{ against } \frac{\tau-\tau_0}{G}$$

on a logarithmic scale; τ_0 was estimated from Equations 5 and 6. Figs. 5 and 6 illustrate this form of plot for Zn–22% Al and Pb–62% Sn, respectively. For Zn–22% Al, Fig. 5 shows that all of the experimental data, which span almost six orders of magnitude of strain rate, cluster about a single straight line with a slope of 2.3. A similar trend is shown for Pb–62% Sn in Fig. 6, where the experimental data fall very close to a straight line having a slope of 1.77. Based on the plots presented in Figs. 5 and 6, the superplastic behaviour of Zn–22% Al and Pb–62% Sn can be represented by

$$\frac{\dot{\gamma}kT}{D_{gb}Gb} = 2.25 \times 10^{6}\left(\frac{b}{d}\right)^{2.3}\left(\frac{\tau-\tau_0}{G}\right)^{2.3}, \quad (7)$$

and

$$\frac{\dot{\gamma}kT}{D_{gb}Gb} = 1.3 \times 10^{5}\left(\frac{b}{d}\right)^{2.4}\left(\frac{\tau-\tau_0}{G}\right)^{1.77}, \quad (8)$$

respectively.

The final step in the present analysis was to check whether Equations 7 and 8 can explain the values of the activation energy measured during superplastic flow at low stresses. Fig. 7, where $\log \dot{\gamma}$ is plotted against $1/T$ for three different stresses, provides a comparison between experimental data of Zn–22% Al which were taken from Mohamed *et al.* [3] and the prediction of Equation 7 which is represented by solid lines. It is clear that, for the range of testing temperatures used, the correspondence between Equation 7 and experimental data is good and that Equation 7, like experimental data, yields an activation energy which increases with decreasing stress. Of course, Equation 7 results in a curved relationship between $\log \dot{\gamma}$ and $1/T$ but, because of experimental scatter and because of a narrow range of temperature used in experiments, it is difficult to distinguish the difference between a fit of the data using a straight line and that provided by Equation 7. Similarly, Fig. 8 gives good agreement between the experimental data and the prediction of Equation 8 for Pb–62% Sn, with one apparent exception: the data obtained from the procedure of temperature cycling (represented by solid diamonds) fall consistently below the prediction of Equation 8 for $\tau = 5.6 \times 10^{-2}$ MPa, although both experiment and Equation 8 give the same

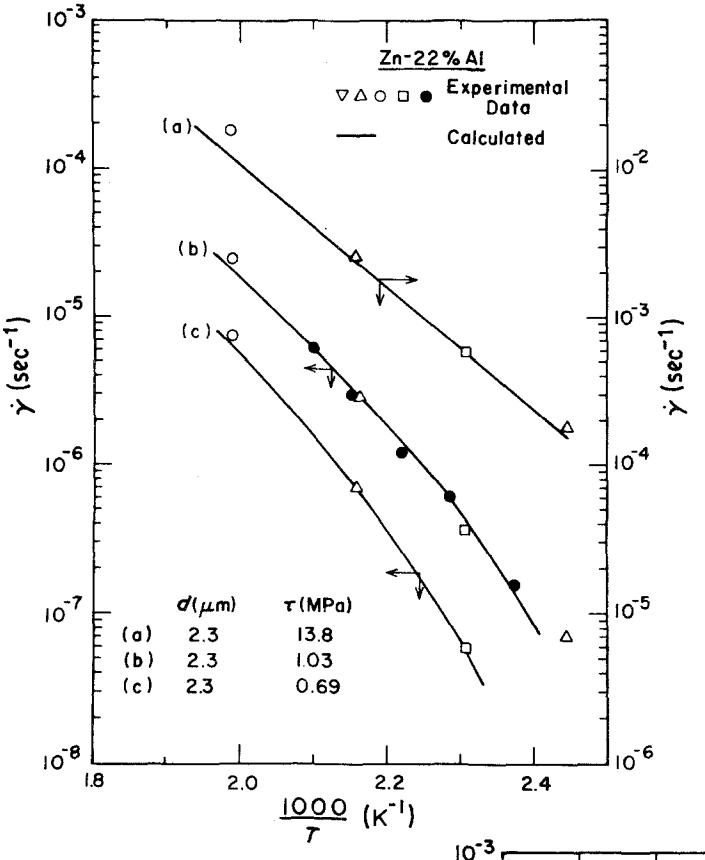

Figure 7 A plot of log of the strain rate against $1/T$ for Zn–22% Al. Data were taken from [3]. Solid circles represent data from temperature cycling procedure.

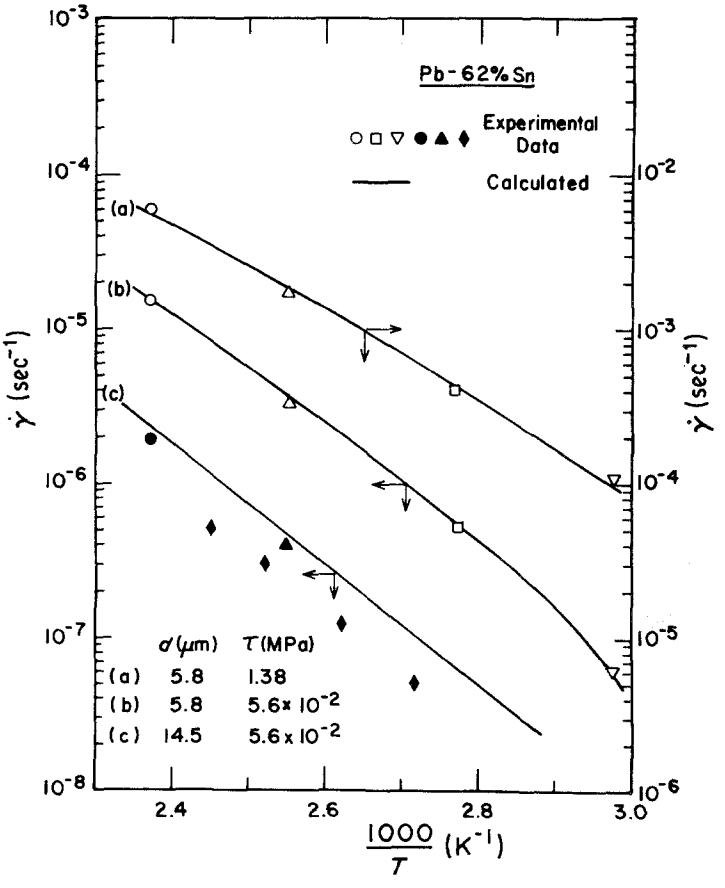

Figure 8 Dependence of log strain rate on $1/T$ for Pb–62% Sn. Data were taken from [4]. Solid diamonds represent data from temperature cylcing procedure.

activation energy. Examination of Figs. 3 and 6 of Mohamed and Langdon [4], from which the experimental data of Pb–62% Sn were taken, suggests that the discrepancy in position between experimental results of temperature cycling and the prediction of Equation 8 is not related to Equation 8 but arises from the fact that the creep rates exhibited by the tested specimen having $d = 14.5\,\mu m$ as a result of sudden changes in temperature were generally slower than those given by $\dot{\gamma} - \tau$ plots for the same grain size.

3. Discussion

The present analysis suggests that the data of Zn–22% Al and Pb–62% Sn in both regions II and I are not necessarily associated with two different deformation mechanisms, since the same data can be equally correlated with a single deformation process that incorporates a threshold stress, τ_0. This correlation is manifested by three observations. First, all of the experimental data obtained for each alloy in both regions, when plotted as normalized creep rates against $(\tau - \tau_0)/G$ on a logarithmic scale, can be fitted with a single straight line. Second, this fit appears as good as that based on the possibility of the operation of two different, but sequential, processes in regions II and I [16]. Third, there is consistency between the Arrhenius plot of $\log \dot{\gamma} - 1/T$ constructed from the rate equations involving τ_0 (Equations 7 and 8) and that obtained experimentally; both plots show an increase in the apparent activation energy with decreasing stress.

The interpretation of the experimental data of Zn–22% Al and Pb–62% Sn at low stresses in terms of a threshold stress implies that regions II and I are controlled by the same deformation mechanism and that the apparent differences in behaviour between the two regions arise from the increasing significance of τ_0 in region I.

It is generally accepted that grain-boundary sliding (GBS) plays the dominant role in the superplastic deformation process. Several theories of superplasticity based on GBS with accommodation by dislocation motion were developed [15, 18, 19, 24, 25] and all of them, despite differences in assumptions and details, lead to rate equations which are consistent with the mechanical characteristics of region II; $n = 2$, $Q = Q_{gb}$ and $s = 2$. Of these theories, the models of Gifkins [18, 19] and Gittus [15] consider that sliding is due to the motion of appropriate boundary dislocations along the boundary plane; the theory of Gifkins considers the movement of grain-boundary dislocations along grain boundaries whereas the theory of Gittus considers the glide of interphase-boundary dislocations* in the interphase boundaries. Because of their considerations that GBS involves the movement of boundary dislocations, the theories of Gifkins [18, 19] and Gittus [15] may provide a possible interpretation of regions II and I in terms of a single deformation process which is associated with a threshold stress.

Boundary dislocations, like lattice dislocations, may interact with various defects and as a result a force will be required to overcome this interaction before dislocations can glide and produce sliding. Depending on the nature of defects, two possible sources of τ_0 were identified [15, 26]. The first source is associated with the interaction between boundary dislocations and features of boundary structure. The pinning of boundary dislocation by ledges provides an example for such an interaction, which was discussed by Gittus [15] and which leads to the following equation:

$$\tau_0 = \frac{E}{b_b L} \tag{9}$$

where E is the energy of the boundary defect per unit length, b_b is the Burgers vector of boundary superdislocation, and L is the width of the broad face of a ledge. The second source is related to the interaction between boundary dislocations and precipitates. According to an analysis by Ashby [26], τ_0 in this case is given by an equation similar to Equation 9 with L equal to the spacing between the precipitate particles. However, a common feature associated with the interaction of boundary dislocations with ledges [15] and precipitates [26] is that τ_0 predicted from such processes, as illustrated by Equation 9, exhibits a very weak temperature dependence†. Accordingly, it is clear that these two sources of τ_0 can-

*Dislocations glide in pairs so that the antiphase boundary created by the passage of the leading dislocation is then erased by the passage of the second dislocation.

†Since E depends on the shear modulus, the temperature dependence of τ_0 is expected to be the same as that of the shear modulus.

not explain Equation 4 which, according to the present analysis, represents the requirement for the interpretation of the experimental data of Zn–22% Al and Pb–62% Sn in terms of a threshold stress process, and that other possible sources of τ_0 need to be explored.

It is quite possible, in view of the conditions under which superplastic alloys are prepared and tested, that impurity atoms segregate at boundaries. Because of the elastic field of the boundary dislocation, which is identical to that for a dislocation of the same Burgers vector in an isotropic elastic continuum [27], segregation would occur preferentially at boundary dislocations [28]. This, in turn, results in locking the dislocation. Under the conditions of strong binding between impurity atoms and dislocations and very low mobility of impurity atoms, the production of appreciable sliding rates may require the breakaway of boundary dislocations from the impurity atmosphere. If it is assumed that λ is the distance between impurity atoms on the dislocation line, the stress, τ_0, required to separate a boundary dislocation from impurity atoms may, as a first approximation*, be given by [29]

$$\tau_0 b_b \lambda \simeq \frac{W}{b} \tag{10}$$

where W is the binding energy between an impurity atom and a boundary dislocation. Additionally, if it assumed that the distribution of impurity atoms around the dislocation as a function of temperature is approximately similar to that of impurity atoms around a lattice dislocation, the following equation can be introduced [29]:

$$\lambda \simeq \frac{b_b}{c_0} \exp - \left(\frac{|W|}{kT}\right) \tag{11}$$

where c_0 is the solute concentration corresponding to $W = 0$. By combining Equations 10 and 11, the normalized threshold stress, τ_0/G, resulting from the pinning of boundary dislocations by impurity atoms may be approximately given by

$$\frac{\tau_0}{G} = B \exp\left(\frac{|W|}{kT}\right), \tag{12}$$

where

$$B = \frac{c_0 W}{b b_b^2 G}.$$

Equation 12 exhibits an interesting feature: the temperature dependence of τ_0 is essentially identical in form to that of the empirical equations based on the analysis of the experimental data (Equations 5 and 6). This implies that $Q = 2.38 \times 10^4$ J mol^{-1} (0.25 eV) and $Q = 1.65 \times 10^4$ J mol^{-1} (0.17 eV) represent the binding energies, W, between an impurity atom and the boundary dislocation, in Zn–22% Al and Pb–62% Sn, respectively. While no documented values for the binding energies between impurity atoms and boundary dislocations are available, the present values of Q are not unreasonable when compared with those estimated from the interaction of impurity atoms with lattice dislocations [29]. Further more, by taking $W = Q$ and substituting approximate values for c_0, b_b and G into the expression

$$B = \frac{c_0 W}{b b_b^2 G},$$

the estimated value of B appears comparable to those of Equations 5 and 6. For example, the value of B for Zn–22% Al was estimated as 8×10^{-8} when W c_0, G, and b_b were taken as 4×10^{-20} J (0.25 eV), 5×10^{-6}, 3.5×10^4 MPa, $b_b = 1.56b$ [27] with $b = 3 \times 10^{-8}$ cm, respectively; according to Equation 5, $B = 2 \times 10^{-8}$. Therefore, it appears that an appoximate analysis based on the pinning of boundary dislocations by impurity atoms can lead to a threshold stress which varies strongly with temperature and which accounts for Equation 4.

Another possible source of τ_0 which is also based on the interaction between impurity atoms and boundary dislocations may be linked to the process of the viscous drag of impurity atoms; this situation is favourable when impurity atoms are very mobile and can catch up with dislocations. Under viscous drag conditions, it is possible, as suggested by Weertman [30], that several viscous processes operate and that one of these processes controls the motion of dislocations while another process produces a frictional stress, τ_0; for example, the Cottrell–Jawson microcreep mechanism [31] could control the motion of dislocations whereas the stress-induced order mechanism [32] might give rise to a temperature dependent frictional stress, τ_0, whose value is proportional to $Gc_0^2 \exp(2W/kT)$. Although the suggestion of Weertman

*Thermal activation is ignored.

[30] deals basically with the interaction between lattice dislocations and impurity atoms, the same concept may still be applicable to the motion of boundary dislocations when opposed by viscous drag processes.

The concept of attributing τ_0 to the pinning of boundary dislocations by impurity atoms leads to an interesting implication*: if, due to a very high purity level of the material or due to some other factors, segregation of impurity atoms at boundaries did not occur, region I ($n > 2$ and $Q > Q_{bb}$) would not be observed; instead, region II with $n = 2$ and $Q = Q_{gb}$ would dominate the whole range of strain rates except the very low strain rates where diffusional creep [33–36] might emerge as a controlling mechanism.

4. Conclusions

By examining the experimental data of Zn–22% Al and Pb–62% Sn, it is shown that the transition between region II (intermediate stresses) and region I (low stresses) can be described by a threshold stress process which depends strongly on temperature. This implies that the sigmoidal behaviour of superplastic alloys at low stresses may not necessarily arise from the operation of two sequential mechanisms. A possible origin of a threshold stress that decreases strongly with increasing temperature is impurity atom segregation at grain boundaries and their interaction with boundary dislocations.

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*Another implication is that the apparent stress exponent and the apparent activation energy in region I would be sensitive to the type and level of impurities that may segregate at boundaries.