Letters

Observations on the magnitude of grain boundary sliding in Region 1 of superplasticity

Much confusion has arisen concerning the relationship between stress and strain rate in superplastic metals. The problem may be appreciated by reference to Fig. 1, which shows a schematic logarithmic plot of steady-state stress, σ, against the imposed strain rate, $\dot{\epsilon}$, for the superplastic Zn–22 wt% Al eutectoid alloy [1].

In the early work of Vaidya *et al.* [2], the experimental points lay along a curve which divided readily into three regions, designated I, II and III, such that the strain rate sensitivity, m $(= \partial \ln \sigma / \partial \ln \dot{\epsilon})$, increased from ~ 0.27 in Region III to ~ 0.5 in Region II and to ~ 1.1 in Region I. Subsequent experiments by Mohamed and Langdon [3] tended to confirm the trends in Regions III and II, but at the lowest strain rates in region I there was a decrease in m to ~ 0.24 so that the results then lay along a sigmoidal curve.

At the present time, the published data for Zn–22 wt% Al divide into two categories in Region I: whereas the results of Vaidya *et al.* [2] were confirmed in later experiments by Misro and Mukherjee [4] and Arieli *et al.* [5], the trend reported by Mohamed and Langdon [3] was subsequently confirmed by Grivas [6] and Vale *et al.* [7]. Furthermore, all six sets of experiments were conducted under similar conditions of stress, temperature and grain size, so that there is no obvious reason for this apparent discrepancy.

It has been demonstrated that, for superplastic metals exhibiting a decrease in m in Region I, there is a corresponding decrease in the measured contribution of boundary sliding to the total strain [8] and also a decrease in the total elongation to failure [9, 10]. However, Arieli and Mukherjee [11] argued recently that these results may be only apparent due to the presence of concurrent grain growth during the tests. The purpose of this note is to demonstrate that this argument is erroneous, and to reiterate that there is a genuine decrease in the sliding contribution in Region I when $m \simeq 0.3$.

Many experiments have been conducted to measure the contribution of sliding to the total strain in one or more of the three regions associated with superplastic metals. The results are summarized in Table I, where ϵ_{gbs} and ϵ_t are the strain due to grain boundary sliding and the total strain, respectively, and the results are for metals exhibiting a sigmoidal relationship between σ and $\dot{\epsilon}$. An examination of these data shows that the calculated values of $\epsilon_{gbs}/\epsilon_t$ are remarkably consistent in Region II, despite the use of several different measuring procedures, and it is clear that $\epsilon_{gbs}/\epsilon_t \simeq 50$ to 70% in this region. It is also apparent from this tabulation that there is a sharp decrease in the values of $\epsilon_{gbs}/\epsilon_t$ in both Regions I and III.

Arieli and Mukherjee [11] concentrated on the results reported earlier by Vastava and Langdon [18] using the Pb–62 wt% Sn eutectic. These results are summarized in Table II, where $\bar{w}$ is the average value of the transverse offsets in a set of longitudinal marker lines at a total strain of 22% in each region, and $\epsilon_{gbs}/\epsilon_t$ gives the estimated percentage strain due to sliding.

The sliding contribution was calculated in this work from the expression

$$\epsilon_{gbs} = \phi \bar{w} / \bar{L} \qquad (1)$$

where ϕ is a geometric constant which was put equal to 1.5 [24] and $\bar{L}$ is the grain size determined from the mean linear intercept along a longitudinal marker.

According to Arieli and Mukherjee [11], "when measurements are made on the *polished* surface of the specimen following the deformation, *$\bar{L}$ is the final grain intercept value*" (emphasis added). These authors then selected hypothetical numbers for the extent of grain growth in Pb–62 wt% Sn, and concluded that the calculated value of $\epsilon_{gbs}/\epsilon_t$ in Region I may underestimate the true value by a factor of 2.5. Reference to Table II shows that an underestimation by this amount would conveniently bring the values of $\epsilon_{gbs}/\epsilon_t$ in Regions I and II into perfect agreement.

In fact, the preceding verbatim statement suggests that Arieli and Mukherjee [11] have

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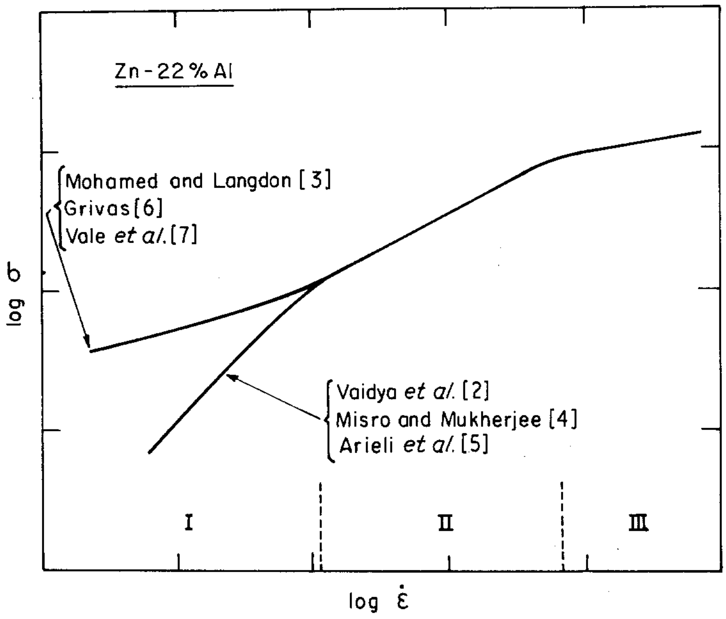

Figure 1 Schematic illustration of stress against strain rate in Zn–22 wt% Al, showing the two types of behaviour reported in Region I.

not appreciated the procedure used for measuring boundary sliding. Four specific points may be noted:

(1) The statement seems to imply that the specimen is polished after deformation and prior to taking the sliding measurements. This is, of course, an impossibility, because the measurements are based on the offsets in marker lines which are placed on the polished surface *before* testing.

(2) The value of $\bar{L}$ in Equation 1 is determined *prior* to the test, and thus it relates unambiguously to the *initial* mean linear intercept.*

(3) If there was major grain growth (much less than the factor of 2.5 assumed by Arieli and Mukherjee [11]), this would tend to obscure the surface marker lines and make it impossible to take the detailed measurements of the transverse offsets. In any case, it has been demonstrated with photomicrographs that there is no significant grain growth in Region I at the selected strain of 22% (Fig. 1 of Vastava and Langdon [18]) and also at the much higher strain of 70% [26].

(4) The argument concerning the influence of grain growth on $\bar{L}$ is superfluous. The implication presented by Arieli and Mukherjee [11] is that the values of $\bar{w}$ are similar in Regions I and II, but that the lower estimate of $\epsilon_{\mathrm{gbs}}/\epsilon_{\mathrm{t}}$ in Region I is due solely to grain growth and a consequent increase in $\bar{L}$. In fact, it is clear from the measurements shown in Table II that the true magnitude of sliding, as represented by $\bar{w}$, decreases sharply in Region I. It is also obvious that the values of $\epsilon_{\mathrm{gbs}}/\epsilon_{\mathrm{t}}$ were calculated for Pb–62 wt% Sn using the same value of $\bar{L}$ in each of the three regions.

As noted earlier, there is a corresponding decrease in the elongation to fracture in Region I [9, 10]. Arieli and Mukherjee [11] also attribute this result to concurrent grain growth by postulating that the accommodation of sliding is more difficult in the presence of grain growth, thereby leading to the nucleation and growth of interfacial voids. Again, this statement is not reasonable.

First, the presence of concurrent grain growth does not lead, *a priori*, to a decrease in fracture strain. This is illustrated in Fig. 2, which shows the elongation to failure as a function of imposed strain rate for specimens of two superplastic materials: Zn–22 wt% Al [10] and Al–33 wt% Cu [27]. For the former material, tested at an absolute temperature, T, of 503 K and with an initial spatial grain size, d_0, of 2.5 μm, there is a genuine Region I with a decrease in m at low strain rates and, since the overall ductility is essentially

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*In the unlikely event that $\bar{L}$ is measured *after* testing, it is necessary to include an additional factor in Equation 1 even in the absence of grain growth [25].

TABLE I Measurements of grain boundary sliding in regions I, II and III of superplasticity

Material	$\epsilon_{gbs}/\epsilon_t$(%)			Reference
	Region I	Region II	Region III	
Al–33 wt% Cu	–	~ 70	–	Hori *et al.* [12]
Al–9 wt% Zn–1 wt% Mg	42	63	26	Matsuki *et al.* [13]
Al–11 wt% Zn–1 wt% Mg	~ 60	~ 80	~ 50	Matsuki *et al.* [14]
Mg–33 wt% Al	12	64	29	Lee [15]
Mg–1.5 wt% Mn	33 ± 4	49 ± 6	30 ± 4	Valiev and Kaibyshev [16]
Pb–62 wt% Sn	–	~ 70	–	Dingley [17]
Pb–62 wt% Sn	21 ± 5	56 ± 12	20 ± 4	Vastava and Langdon [18]
Pb–62 wt% Sn	–	50	–	Furushiro and Hori [19]
Zn–0.4 wt% Al	~ 40	~ 50	~ 30	Kaibyshev *et al.* [20]
Zn–0.4 wt% Al	$\leqslant$ 30	> 50	Decreases	Kaibyshev *et al.* [21]
Zn–22 wt% Al	~ 30	~ 60	~ 30	Holt [22]
Zn–22 wt% Al	< 30	~ 60	< 20	Novikov *et al.* [23]

proportional to the strain rate sensitivity [28], there is a corresponding decrease in the elongation to fracture in this region. As indicated in Fig. 2 by the values of the spatial grain size, d, recorded at fracture, the extent of concurrent grain growth is very small in Zn–22 wt% Al, and does not exceed a factor of two (from 2.5 to 5.0 μm) at the very slowest strain rate. By contrast, Al–33 wt% Cu exhibits very extensive grain growth during testing, such that the mechanical measurements give rise to a false "Region I" with an apparent low value of m at the slowest strain rates [27]. However, there is no corresponding diminution in the elongation to fracture at these slow rates, *even though the grains grow in size by a factor of thirteen* (from 1.5 to 20 μm).

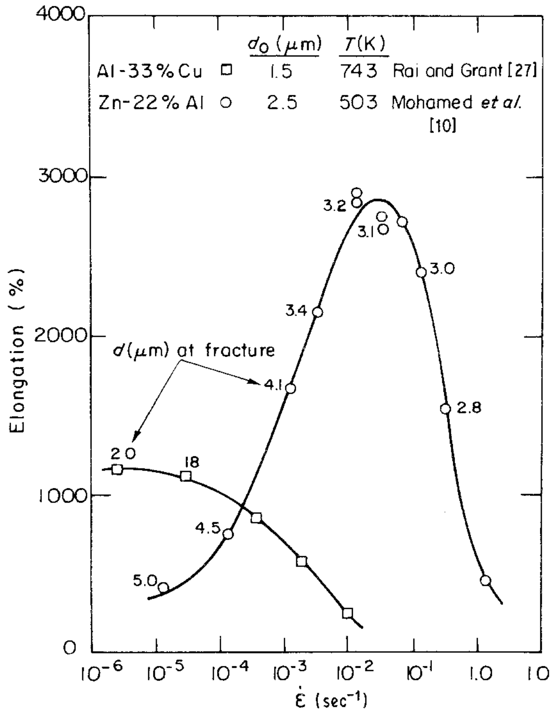

Figure 2 Elongation at failure versus initial strain rate for Al–33 wt% Cu and Zn–22 wt% Al, showing the measured grain size at fracture.

In practice, therefore, the two sets of experimental results shown in Fig. 2 are mutually consistent [29], because Zn–22 wt% Al exhibits a genuine Region I with low strain rate sensitivity and low elongations at the slow strain rates whereas Al–33 wt% Cu does not exhibit a genuine Region I, at least over the strain rates covered experimentally, and, despite massive grain growth, there is no corresponding decrease in the elongation to failure.

Second, failure occurs in Zn–22 wt% Al in region I not through the growth and catastrophic interlinkage of cavities but because of the formation and development of macroscopic necking [30]. Again, this observation is consistent with a true decrease in the value of m in this region.

Finally, it should be noted that, whereas it has been demonstrated [1, 29] that the suggestion of concurrent grain growth is not able to account for the low value of m obtained in some experi-

TABLE II Measurements of sliding in Pb–62 wt% Sn [18]

Region	$\bar{w}$(μm)	$\epsilon_{gbs}/\epsilon_t$(%)
I	0.30	21 ± 5
II	0.79	56 ± 12
III	0.28	20 ± 4

ments [3, 6, 7] on Zn–22 wt% Al in Region I, there is good evidence that the apparent high values of m reported in some other experiments [2, 4, 5], as indicated in Fig. 1, are due to a failure to accurately take into account the primary stage of creep. This is discussed in more detail elsewhere [1].

In summary: (1) there is a genuine decrease in the contribution of grain boundary sliding at low strain rates in Region I of superplasticity, and this decrease cannot be attributed to the occurrence of concurrent grain growth; (2) similarly, the decrease in elongation in Region I is not due to concurrent grain growth but to the formation and development of macroscopic necking.

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References

1. D. GRIVAS, J. W. MORRIS and T. G. LANGDON, *Scripta Met.* **15** (1981) 229.
2. M. L. VAIDYA, K. L. MURTY and J. E. DORN, *Acta Met.* **21** (1973) 1615.
3. F. A. MOHAMED and T. G. LANGDON, *ibid.* **23** (1975) 117.
4. S. C. MISRO and A. K. MUKHERJEE in "Rate Processes in Plastic Deformation of Materials," edited by J. C. M. Li and A. K. Mukherjee (American Society for Metals, Metals Park, Ohio, 1975) p. 434.
5. A. ARIELI, A. K. S. YU and A. K. MUKHERJEE, *Met. Trans. A* **11A** (1980) 181.
6. D. GRIVAS, Report No. LBL-7375 (Lawrence Berkeley Laboratory, University of California, Berkeley, 1978).
7. S. H. VALE, D. J. EASTGATE and P. M. HAZZLEDINE, *Scripta Met.* **13** (1979) 1157.
8. D. M. R. TAPLIN, G. L. DUNLOP and T. G. LANGDON, *Ann. Rev. Mater. Sci.* 9 (1979). 151.
9. H. ISHIKAWA, F. A. MOHAMED and T. G. LANGDON, *Phil. Mag.* **32** (1975) 1269.
10. F. A. MOHAMED, M. M. I. AHMED and T. G. LANGDON, *Met. Trans. A* **8A** (1977) 933.
11. A. ARIELI and A. K. MUKHERJEE, *J. Mater. Sci.* **15** (1980) 2401.
12. S. HORI, N. FURUSHIRO and S. KAWAGUCHI, Proceedings of the 19th Japan Congress on Materials Research (The Society of Materials Science, Kyoto, Japan, 1976) p. 1.
13. K. MATSUKI, H. MORITA, M. YAMADA and Y. MURAKAMI, *Met. Sci.* **11** (1977) 156.
14. K. MATSUKI, Y. UENO and M. YAMADA, *J. Jap. Inst. Met.* **38** (1974) 219.
15. D. LEE, *Acta Met.* **17** (1969) 1057.
16. R. Z. VALIEV and O. A. KAIBYSHEV, *Phys. Stat. Sol.* (*a*) **44** (1977) 65.
17. D. J. DINGLEY, "Scanning Electron Microscopy/1970" (IIT Research Institute, Chicago, Illinois, 1970) p. 329.
18. R. B. VASTAVA and T. G. LANGDON, *Acta. Met.* **27** (1979) 251.
19. N. FURUSHIRO and S. HORI, *Scripta Met.* **13** (1979) 653.
20. O. A. KAIBYSHEV, R. Z. VALIEV and V. V. ASTANIN, *Phys. Stat. Sol.* (*a*) **35** (1976) 403.
21. O. A. KAIBYSHEV, SH. KH. KHANNANOV and R. Z. VALIEV, Proceedings of the 4th International Conference on the Strength of Metals and Alloys, Vol. 1 (Laboratoire de Physique du Solide, ENSMIM, Nancy, France, 1976) p. 423.
22. D. L. HOLT, *Trans. AIME* **242** (1968) 25.
23. I. I. NOVIKOV, V. K. PORTNOY and T. E. TERENTIEVA, *Acta Met.* **25** (1977) 1139.
24. T. G. LANGDON, *Met. Trans.* **3** (1972) 797.
25. A. GITTINS, *Trans. AIME* **239** (1967) 922.
26. D. A. MILLER, R. B. VASTAVA and T. G. LANGDON, "Microstructural Science", Vol. 9 (Elsevier North Holland, New York) (in press).
27. G. RAI and N. J. GRANT, *Met. Trans. A* **6A** (1975) 385.
28. T. G. LANGDON, *Scripta Met.* **11** (1977) 997.
29. *Idem*, in "Recrystallization and Grain Growth of Multi-Phase and Particle Containing Materials", edited by N. Hansen, A. R. Jones and T. Leffers (Risø National Laboratory, Roskilde, Denmark, 1980) p. 223.
30. M. M. I. AHMED, F. A. MOHAMED and T. G. LANGDON, *J. Mater. Sci.* **14** (1979) 2913.

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