determines these kinetics. Boksay's equations (Equations 6–8 in [1]) are valid in the liquid double layer, but cannot be applied to the surface region of the solid unless we assume that a hydrogen-ion defect coexists with the interstitial alkali in the solid. Then Equation 8 of [1] would be invalidated because the separate ion fluxes would be in the same direction, and generally not equal in magnitude unless their defect densities, activation energies and diffusion coefficients were fortuitously equal.

Boksay did not comment on the necessity for any model proposed to achieve quantitative agreement with experimental results. The model we have proposed yields reasonable "numbers" without invoking a variety of other contributing factors. Adherence to the diffusion model, however, produces difficulties in this respect. One such problem is the observation that the diffusion coefficient of alkali ions in the surface region of a glass can be three orders of magnitude larger than that for the bulk. Diffusion model proponents attribute this to the presence of a "swollen" surface layer, in which, ostensibly, the barrier to diffusion is lowered. From simple diffusion theory [9] one can calculate that the lattice would have to "swell" by a factor of more than thirty in order to account for such an increase in diffusion coefficient. On the other hand, the large electric fields present in the surface region can readily induce ion drift velocities ($v = \mu E$) corresponding to those observed experimentally.

Models are useful in that they allow us to conceptualize and test hypotheses related to physical phenomena. However, they cannot be held inviolate when they cease to provide the right answers. Professor Boksay quotes Aristotle to the effect that a true conclusion may be drawn from a false premise. With respect to the simple diffusion model, we believe, following Euripides (Aeolus, fragment 32), that a bad beginning makes a bad ending.

References

1. Z. BOKSAY, *J. Mater. Sci.* 15 (1980) 2398.
2. W. M. MULARIE, W. F. FURTH and A. R. C. WESTWOOD, *J. Mater. Sci.* 14 (1979) 2659.
3. H. N. STEIN, *Adv. Coll. Int. Sci.* 11 (1979) 76, 95.
4. R. J. CHARLES, *J. Electrochem. Soc.* 116 (1969) 1514.
5. Z. BOKSAY, G. BOUQUET and S. DOBOS, *Phys. Chem. Glasses* 9 (1968) 69.
6. R. W. DOUGLAS and T. M. M. EL-SHAMY, *J. Amer. Ceram. Soc.* 50 (1967) 1.
7. J. M. HORN, Jr. and G. Y. ONODA, Jr., *J. Amer. Ceram. Soc.* 61 (1978) 523.
8. A. MANY, Y. GOLDSTEIN and N. B. GROVER, "Semiconductor Surfaces" (North Holland Publishing Company, Amsterdam, 1965) Ch. 4.
9. B. I. BOLTAKS, "Diffusion in Semiconductors" (Academic Press, New York and London, 1963) ch. II.

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W. M. MULARIE
*Central Research Laboratories,
3M Company,
St. Paul, Minnesota, 55101, USA*
W. F. FURTH
A. R. C. WESTWOOD
*Martin Marietta Laboratories,
Baltimore, Maryland, 21227, USA*

On the region I in superplasticity

In several investigations of superplasticity at low strain rates, a region (region I) characterized by high n value and diffusivity equal to that for lattice self-diffusion, was reported [1–3]. The deformation in region I is accompanied by lower elongation to fracture [4, 5] and grain boundary sliding contribution to the total strain [6] than in region II.

Microstructural investigations reveal [7] that all the features characteristic to region II are also present during deformation in region I with the addition that more grain growth occurs. More grain growth in this region is understandable considering the longer test times involved at these low strain rates.

Arieli and Mukherjee [8] have shown for a Zn–22% Al eutectoid alloy that the high n and diffusivity values in region I are only apparent

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and are due to the concurrent grain growth during the superplastic flow. Concurrent grain growth during superplastic deformation also explains the lower elongations to fracture obtained in this region. According to a recent model for the rate-controlling mechanism in superplasticity [9], when concurrent grain growth occurs, the accommodation mechanism for grain boundary sliding gradually loses its effectiveness and voids nucleate and grow at the interfaces leading to early fracture. An indirect and circumstantial proof that the elongation to fracture in region I is controlled by the concurrent grain growth is provided by Mohamed *et al.* [5], who found that in region I for constant strain-rate and temperature the coarser the initial grain size the larger the elongation to fracture, whereas in region II this trend is reversed. These observations are compatible with another observation [10] that the rate of concurrent grain growth is higher for the smaller initial grain sizes and that more grain growth occurs at low strain rates [7], whereas in region II the effect of concurrent grain growth is less important due to much shorter testing times.

The lower contribution of grain boundary sliding to the total strain in this region can also be explained by taking into account the effect of the concurrent grain growth. Usually, the sliding contribution to the total strain is calculated using the relation [6]

$$\epsilon_{\text{gbs}} = \text{k}\bar{\omega}/\bar{L} \qquad (1)$$

where ϵ_{gbs} is the strain due to grain boundary sliding, k is a constant, $\bar{\omega}$ is the average displacement from all boundaries and is measured from line marker experiments, and $\bar{L}$ is the average grain intercept. When measurements are made on the polished surface of the specimen following the deformation, $\bar{L}$ is the final grain intercept value. But the grain boundary sliding took place continuously and, consequently, $\bar{\omega}$ is the displacement at the boundaries for a whole range of grain intercepts between the initial value $\bar{L}_0$ and the final value $\bar{L}$. Since the grains grow according to a power law [10], the average grain intercept, $\bar{L}_{\text{av}}$, value will probably be closer to the initial grain intercept value, $\bar{L}_0$, than the final grain intercept value, $\bar{L}$. Assuming that the ratio $\bar{L}/\bar{L}_0$ will vary between 1.5 and 4 [10], with an average value of 3, then using $\bar{L}_{\text{av}} \simeq \bar{L}/2.5$ in Equation 1 instead of $\bar{L}$ we obtain

$$\frac{\epsilon_{\text{gbs}}(\bar{L}_{\text{av}})}{\epsilon_{\text{gbs}}(\bar{L})} \simeq \frac{\bar{L}}{\bar{L}_{\text{ave}}} \simeq 2.5 \qquad (2)$$

where $\epsilon_{\text{gbs}}(\bar{L}_{\text{av}})$ is the contribution of grain boundary sliding to the total strain when the average grain intercept, $\bar{L}_{\text{av}}$, is used and, $\epsilon_{\text{gbs}}(\bar{L})$ is that obtained when the final grain intercept, $\bar{L}$, is used.

This suggests that it is quite possible that some of the characteristics of the region I, i.e. high n-value, high diffusivity value, lower elongation to fracture and contribution of the grain boundary sliding to the total strain, than in region II, may only be apparent, arising because of concurrent grain growth during superplastic flow.

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References

1. F. A. MOHAMED and T. G. LANGDON, *Acta. Met.* 23 (1975) 117.
2. *Idem, Phil. Mag.* 32 (1975) 697.
3. S-A. SHEI and T. G. LANGDON, *Acta. Met.* 26 (1978) 639.
4. M. M. I. AHMED and T. G. LANGDON, *Met. Trans.* 8A (1977) 1832.
5. F. A. MOHAMED, M. M. I. AHMED and T. G. LANGDON, *ibid.* 8A (1977) 933.
6. B. VASTAVA and T. G. LANGDON, *Acta. Met.* 27 (1979) 251.
7. A. ARIELI and A. K. MUKHERJEE, *Mat. Sci. Eng.* to be published.
8. *Idem, Scripta Met.* 13 (1979) 331.
9. *Idem. Mat. Sci. Eng.* 45 (1980).
10. A. ARIELI, Ph.D. Dissertation, University of California Davis (1979).

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A. ARIELI
A. K. MUKHERJEE
Materials Science Section,
Department of Mechanical Engineering,
University of California,
Davis, CA 95616
USA