JOURNAL OF MATERIALS SCIENCE 12 (1977) 915–918

Cavitation during diffusion creep of a magnesium alloy

E. H. AIGELTINGER
Department of Materials Science, University of Florida, Gainesville, Florida, USA
R. C. GIFKINS
Department of Metallurgy, University of Melbourne, Parkville, Victoria 3052, Australia

Cavities associated with grain-boundary triple edges have been found in a Mg–0.55 Zr alloy after diffusion creep and the angular distribution of cavitated boundaries measured; cavities occur predominantly on boundaries normal to the stress axis and are rounded in form. Nucleation of cavities can, therefore, be associated with grain-boundary sliding and their growth with diffusion. It is suggested that similar cavities found in some alloys after superplastic deformation may also result from significant diffusion creep but this is not typical of all superplastic deformation.

1. Introduction

The presence of cativies during creep has been widely associated [1] with the operation of grain-boundary sliding (GBS). In normal recovery creep (RC) the maximum amount of GBS occurs on grain boundaries lying at about 45° to the stress axis, that is on the boundaries which experience the largest shear stress in a simple tensile specimen [2]. Consequently, the 45° boundaries might be expected to be the most heavily cavitated if GBS controls nucleation or growth of the cavities. On the other hand, the growth of cavities may be controlled by diffusion of vacancies to them; in this case it is the maximum tensile stresses which should be controlling and this leads to the expectation that the 90° boundaries will be the most cavitated.

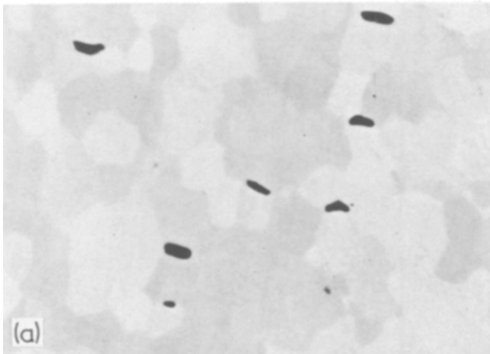
In the context of these remarks it is usual to monitor cavitation by counting the number of boundaries containing one or more cavities in ranges of the angle θ which the grain-boundary trace makes with the stress axis. The measured data have to be corrected to allow for the fact that the trace angle, θ, is not the true angle, θ_r, between the grain-boundary plane and the stress axis. True distribution frequencies, Q_r, may be obtained from Q_a, the frequencies measured for θ, by a treatment due to Scriven and Williams [3] and this has been used here.

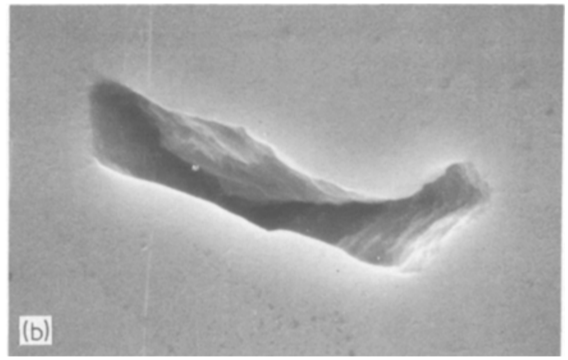
Cavitation has also been observed in association with superplasticity [4], in a limited number of alloys. However, there have not been any reports of it occurring during one of the other principal kind of creep, diffusion creep (DC). The present paper describes cavities observed in a magnesium alloy "Magnox Zr55" (Mg–0.55 Zr) deformed under DC conditions and gives the results of a determination of the angular distribution of boundaries containing cavities.

2. Experimental procedure and results

Details of the sample preparation and testing have been given previously [5]. The present results were obtained on a specimen stressed at 4.57 MPa at 450° C which extended 6.57% at a rate of $3.54 \times 10^{-7} \text{sec}^{-1}$; the creep rate was uniform throughout most of the test, there being a slight acceleration during the last 15% of the test probably as a result of the cavitation which occurred. Other evidence in addition to the linear creep curve has previously been given [5] to show that this specimen deformed by DC.

Cavities were readily seen using the optical microscope on an unetched, polished section (Fig.





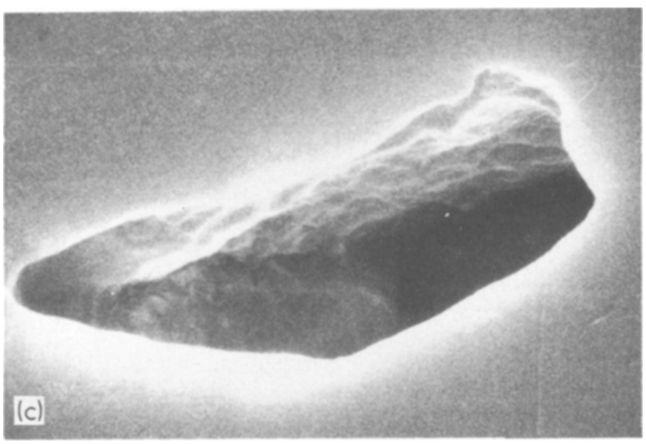


Figure 1 (a) General distribution of cavities, unetched section; stress axis vertical. Optical micrograph, $\times 60$. (b) Cavity at triple edge, unetched. Scanning electron micrograph ($30°$ tilt), $\times 400$. (c) As (b), $\times 400$. (Reduced 75% in reproduction.)

1a). Etching revealed that all cavities were lying on grain boundaries and associated with one, or sometimes two, triple-edge junctions. It proved very difficult to avoid distorting their apparent shape. Insufficient final-stage polishing could lead to fins of material partially blocking the cavity but, on the other hand, prolonged polishing or the slightest trace of moisture on the polishing pad could lead to etching and rounding of the cavity mouth. Repeated polishing and examination of the same areas each time led to the conclusion that the sections of the cavities always had rounded forms. This is best illustrated by scanning electron micrographs such as Fig. 1b and c. Sometimes there appeared to be crack-like extremities to cavities, but tilting the specimen to various angles with respect to the incident beam of electrons showed that this was a parallax effect and that the true section was indeed rounded.

Grain-boundary traces containing cavities were sampled along linear traverses parallel to the stress axis. The angle, θ, between the boundary trace and the stress axis for each boundary having one or more cavity was measured and the values of θ divided into $5°$ classes to give values of Q_a. A total of 250 cavitated boundaries was measured, as recorded in Table I. These results were processed using the procedure developed by Scriven and Williams [3] to obtain values of Q_r from those of Q_a. The frequency in terms of Q_r as a function of θ is shown in Fig. 2.

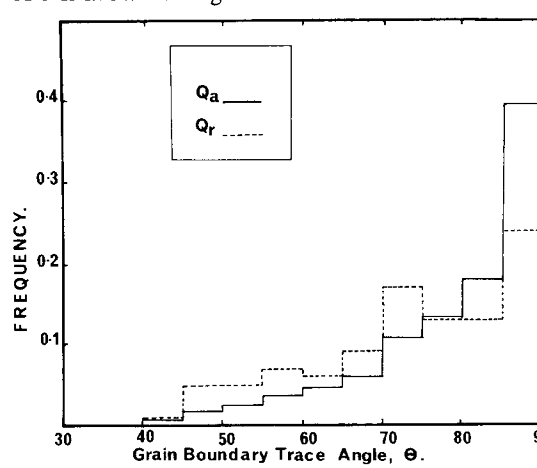


Figure 2 Apparent (Q_a) and real (Q_r) distributions of cavitated boundaries with respect to grain-boundary trace angle θ.

TABLE I Distribution of cavitated boundaries

θ deg	No. of boundaries	Frequency	
		Q_a	Q_r
40–45	1	0.004	0.01
45–50	4	0.016	0.05
50–55	6	0.024	0.05
55–60	9	0.036	0.07
60–65	11	0.044	0.06
65–70	15	0.060	0.09
70–75	27	0.108	0.17
75–80	33	0.132	0.13
80–85	45	0.180	0.13
85–90	99	0.396	0.24

3. Discussion

It is apparent from Fig. 2 that more than $\frac{2}{3}$ of the cavities present lie in grain boundaries in the range $\theta = 70$ to $90°$ and none lie on boundaries for which $\theta < 40°$. This kind of distribution has previously been found to be favoured by large grain sizes [6], low rates of strain [7], or high temperatures [8], and has generally been taken to imply that the dominating process is growth of cavities by diffusion. However, in the previous cases the cavities concerned have been those which occur discretely along the grain boundaries (r-type), whereas in the present case they were all associated with one or two triple lines. The evidence indicates that these cativites nucleate at triple edges, in the manner of w-type cavities, but generally only on boundaries lying approximately normal to the stress axis ($\theta \simeq 90°$). This would be expected from the distribution of sliding boundaries during DC beacuse it has been shown [5] that GBS is then a maximum for grain boundaries which have $\theta \simeq 30°$. On average, each boundary with $\theta \simeq 90°$ has two boundaries with $\theta \simeq 30°$ meeting it (see Fig. 3), and GBS on these will nucleate a cavity as indicated by the broken lines in Fig. 3. Further sliding can cause such cavities to grow, but the maximum diffusional growth should also occur for these cavities which are on boundaries near normal to the stress axis; thus both mechanisms of growth may operate.

The fact that the cavities have rounded shapes and not the typical wedge shape indicates that the diffusional component of growth is very active, if not overriding. This is to be expected when deformation itself is diffusion controlled. It is of interest to note that similar rounded cavities at triple edges have been noted during the superplastic deformation of an α-β brass [9]. Of even greater interest is the suggestion, based on an analysis of the contributions various mechanisms make to the overall deformation [10], that DC is usually important in the superplastic range of deformation for this material (see Fig. 11 of [10]). In contrast, for other superplastic materials, analysis of typical results [10] shows DC only at the very lowest stresses of the superplastic range. Examples of this for several materials are given in [10] (e.g. in Figs. 8 to 10). This offers a possible explanation of the drop in ductility of superplastic materials tested at low stresses (as they enter the low-stress regime I), if DC is more prone to give cavitation than regime II. However, no microstructural evidence for this has been noted so far.

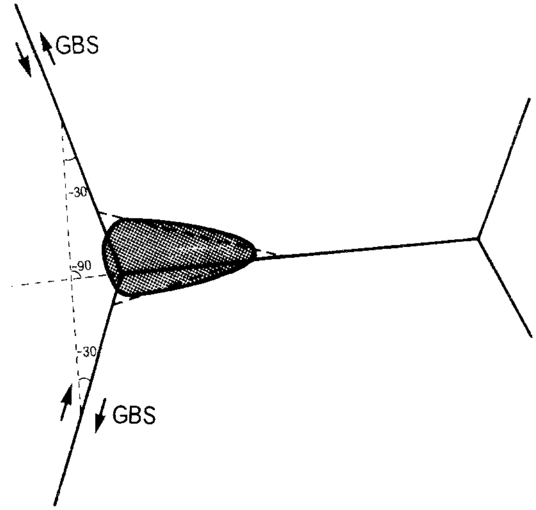


Figure 3 Schematic representation of formation of cavity at triple edge by sliding on two boundaries incident upon $\theta = 90°$ boundary.

It is possible, therefore, to explain the rounded triple-edge cavities in superplastic materials as being due to the operation of a significant amount of DC. Because they are rounded, these cavities do not readily grow as cracks, and because they are nucleated only at specific triple edges they do not readily link to form critical length cracks. Hence, reasonably high ductilities may be preserved even in the presence of cavities. Cavitation may not always occur during DC, because grain growth frequently accompanies DC [5, 11] and could thus eliminate cavity nuclei before they stabilize. It may also be necessary for there to be nucleating particles for cavities even at triple-edges, as recently found in a number of cases [12]; there would be adequate zirconium hydride particles in the present material for this purpose.

If the above suggestions are accepted, the fact that most superplastic materials do not cavitate in the superplastic range is an argument against acceptance of the Ashby and Verrall [13] DC mechanism for superplasticity.

4. Conclusions

Under DC conditions for the material considered here (MG–0.55 Zr), cavities are nucleated at triple edges by GBS and grow dominantly by diffusion under the stresses normal to the cavitated boundaries. It is noted that some cases of cavitation during superplastic deformation may be associated with an unusually large contribution from diffusion creep and that cavitation then apparently occurs in the manner described.

Acknowledgements

This work was performed in the Department of Metallurgy, University of Melbourne whilst one of us (EHA) was in receipt of a Post-doctoral Fellowship financed by the Commonwealth Scientific and Industrial Research Organization; R.C.G. is an officer of the Tribophysics Division of that Organization, seconded to the Metallurgy Department.

References

1. A. J. PERRY, *J. Mater. Sci.* **9** (1974) 1016.
2. R. C. GIFKINS, A. GITTINS, T. G. LANGDON and R. L. BELL, *ibid* **3** (1968) 306.
3. R. A. SCRIVEN and H. D. WILLIAMS, *Trans. Met. Soc. AIME* **233** (1965) 1593.
4. J. W. EDINGTON, K. N. MELTON and C. P. CUTLER, *Prog. Mat. Sci.* **21(2)** (1975) 61.
5. E. H. AIGELTINGER and R. C. GIFKINS, *J. Mater. Sci.* **10** (1975) 1903.
6. D. M. R. TAPLIN, *Phil. Mag.* **20** (1969) 1079.
7. A. GITTINS and H. D. WILLIAMS, *ibid* **16** (1967) 849.
8. R. L. FLECK, G. C. COCKS and D. M. R. TAPLIN, *Met. Trans.* **1** (1970) 3415.
9. S. SAGAT, P. BLENKINSOP and D. M. R. TAPLIN, *J. Inst. Metals* **100** (1972) 268.
10. R. C. GIFKINS, *Met. Trans.* **A7** (1976).
11. B. BURTON and G. W. GREENWOOD, *Met. Sci. J.* **4** (1970) 215.
12. R. L. PLAYER and G. BRINSON, *J. Australian Inst. Metals* **20** (1975) 226.
13. M. F. ASHBY and R. A. VERRALL, *Acta Met.* **21** (1973) 149.

Received 27 September and accepted 14 October 1976.