FRACTURE DURING SUPERPLASTIC FLOW OF INDUSTRIAL Al-Mg ALLOYS

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INTRODUCTION

Superplasticity has now evolved from the province of arcane academic studies and metallurgical legerdemain to the realm of industrial exploitation. Superplastic forming is used in the fabrication of a wide range of components, usually in fairly short production runs, where low tooling costs counter the disadvantage of the necessarily low strain rates. The alloys used industrially usually exhibit superplasticity by virtue of possessing fine, stable microstructures. These are obtained in materials having a microduplex structure or a fine grained recrystallized microstructure ($d \leq 10$ μm) stabilized by well dispersed grain boundary precipitates. The techniques for obtaining suitable microstructures are now well established. Earlier work at Waterloo [1, 2, 3] has demonstrated that a wide range of copper base alloys can readily be rendered superplastic, and potentially useful in forming processes. The work reported here shows that the same is true of industrial Al-Mg alloys.

Edington, Cutler and Melton [4] have published a comprehensive review of superplasticity, but the fracture and cavitation behaviour of superplastic alloys has received close attention only recently [5 - 9]. The cavitation discussed here occurs throughout the gauge length of the specimen and should be distinguished from the local cavitation which may occur close to the fracture surface in a sample which has failed by necking. It is apparent that this type of cavitation commonly accompanies the high temperature flow of Cu, Al, Fe and Ni based materials containing second phase particles: fine grained alloys are no exception. In fact, cavity density generally increases with decreasing grain size, so that these alloys in the superplastic condition frequently fracture in a quasi brittle manner, with minimal necking, even though high elongations are attained. This is due to autocatalytic linking of cavities.

The micromechanisms of fracture in a cavitating superplastic alloy can be considered from different viewpoints. This may produce a fruitful confluence of theoretical approaches, providing a unifying influence in line with the overall objectives of the Conference. Firstly, the fracture process may be studied in terms of well established models of low temperature ductile fracture, where voids or cavities nucleate on second phase particles at relatively low strains and grow and move together by plastic flow until coalescence by shear causes failure. Here the orientation is towards material parameters like n and m. Secondly, the process may be considered as a high temperature intergranular fracture problem. Here voids nucleate and grow at grain boundaries as a consequence of grain boundary sliding and vacancy diffusion, until failure occurs by cavity interlinkage and propagation when a critical level of damage is achieved [10]. Thirdly, failure can be regarded as a manufacturing problem, where incipient fracture

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limits permissible strain levels in forming operations, and cavitation affects mechanical properties of the finished product.

This paper reports initial results from a study of cavitation and superplastic flow in four Al-Mg alloys manufactured by ALCAN. When fine grained, these alloys are moderately superplastic, exhibiting maximum elongations up to 500%. The work was conducted with the aim of providing data on the effects of temperature and strain rate on cavitation in such alloys in order to understand the criteria for ductile fracture and void linkage in strain-rate sensitive alloys.

EXPERIMENTAL

The composition of the alloys investigated is given in Table 1. Alloys A, B and C were experimental melts based on variations of the industrial 5083 composition. The experimental melts were prepared as 12 mm thick castings. They were homogenized at 808 K for 24 hours, then hot rolled to 6 mm. The 5083 was received as 9 mm plate. Final cold rolling of the four alloys was carried out in small increments down to 1.16 mm, with an intermediate anneal to obviate cracking in the case of the thicker 5083, followed by recrystallization at 773 K for 30 minutes. The mean linear intercept grain sizes obtained fell in the range 8 - 12 μm for the four compositions. Flat tensile testpieces of gauge length 13 mm and gauge section 1.16 x 10.0 mm were used. Tests were conducted in a 2.5 kN capacity table model Instron machine with the testpieces enclosed in a furnace regulated to ± 2 K. The strain rates quoted refer to the original gauge length of the testpieces, and most tests were conducted at constant rates of crosshead displacement.

RESULTS

1. Temperature and Strain Rate Effects

Two series of tensile tests were carried out on each of the alloys, A, B and C. First the effect of temperature variation in the range 573 - 843 K was studied at a constant strain rate of $2.6 \times 10^{-6}\ s^{-1}$, which preliminary tests had indicated would be appropriate in promoting superplasticity. The results are given in Table 2, and show the existence of an optimum temperature for superplastic elongation, in agreement with other data, e.g., [7]. The optimum temperatures are listed in Table 2.

The second series of tests was conducted on each alloy at its approximate optimum temperature, in order to determine the effect of strain rate. Results are listed in Table 3 and, as with temperature, the existence of an optimum is indicated, although at least one further test at a low strain rate ($< 6 \times 10^{-5}\ s^{-1}$) is needed to confirm this conclusion for alloy B. This trend is also in accord with published data on the Zn-Al eutectoid [11] and ternary brass [7]. Approximate values of the optimum strain rate for each alloy are given in Table 3. Clearly a properly designed series of factorial experiments would have to be performed to fully optimize test conditions for maximum strain to fracture. However, there is no evidence of marked irregularities on the temperature/strain rate/elongation surface and it is probably that the conditions found are close to the true optimum.

2. Cavitation

A quantitative television microscope (QTM) was used to determine the area fraction of cavities in the region immediately adjacent to the fracture surface in specimens of alloys A and B tested at a range of temperatures and strain rates. The results are listed in Table 4. In both alloys the area fraction of cavities at fracture is a maximum at the conditions for maximum elongation. Neglecting differences in strain rate and test temperature there is a roughly linear relation between area fraction of cavities and elongation, see Figure 1. Thus, these results are clearly in agreement with the previously reported [1, 12, 13] and predictable trend to increasing cavitation with increasing strain. In fact Ridley et al [13] have shown that, in a range of superplastic materials, the rate of cavitation is almost independent of strain, so that representation of the data as in Figure 2 is not an unrealistic exercise, although grain coarsening may have an effect, as described below. Further confirmation that the degree of cavitation with strain is almost independent of strain rate was obtained in a series of tests on alloy 5083. Three series of specimens were strained initially at $6.4 \times 10^{-6}\ s^{-1}$ to a strain of 200% to establish a cavitated structure. They were then strained at $6.4 \times 10^{-6}\ s^{-1}$, $1.3 \times 10^{-2}\ s^{-1}$ and $2.6 \times 10^{-2}\ s^{-1}$ to a range of strains up to fracture so that cavity density could be determined metallographically. No significant effect of strain rate was apparent.

DISCUSSION

The results show that conventional industrial Al-Mg alloys can be readily produced in a superplastic condition. Maximum superplastic elongation was obtained at intermediate values of strain rate and temperature. The relation between applied stress and strain rate for a superplastic alloy is generally represented by:

$$\sigma = A\dot{\varepsilon}^{m} \qquad (1)$$

where A is a constant. It has been shown that maximum ductility is found when the strain rate sensitivity is a maximum. This observation was approximately confirmed by the data obtained on A, B and C and from some further tests on 5083, although sufficient data for accurate calculation of the respective values of m was not obtained.

The fracture process in a cavitating superplastic material probably parallels high temperature creep fracture more closely in its early stages and low temperature ductile fracture as deformation advances. Cavities initiate at second phase particles in grain boundaries by a combination of sliding and diffusion. The strain level at initiation may then be lower than typical levels for low temperature ductile fracture of materials containing a well bonded non-deforming second phase. Recent work has shown that cavitation during tensile superplastic deformation can occur in aluminum bronzes [2, 3], α/β brass [1, 7, 13], nickel silver [9, 13], stainless steel [9, 13] and the Zn-22% Al eutectoid [8, 14]. It is relevant to review briefly the essential features of the cavitation mechanism. Most models of cavitation have been developed to describe behaviour during creep, but there is a close similarity to superplasticity as regards the deformation processes involved, and the models can be related to both processes.

Nucleation of cavities generally arises as a consequence of the blockage of grain boundary sliding at inclusions. In aluminum-bronzes cavities were observed to form at triple junctions and at α/β interfaces [2]. In α/β brasses nucleation sites were identified as α/β interfaces and grain boundary particles [1, 7] and triple junctions [13], with similar observations in the other materials. The effectiveness of hard particles as cavity nuclei was illustrated by Ridley and Livesey [9] in experiments on the Pb-Sn eutectic, which does not normally cavitate during superplastic deformation, but was found to cavitate after the addition of copper and the introduction of a hard intermetallic phase.

The mechanism of nucleation at triple junctions is difficult to justify theoretically during superplastic flow of alloys. The Stroh [15] condition for nucleation is:

$$\tau_c^2 < \frac{12\ \gamma\ G}{\pi L} \tag{2}$$

where L represents either the length of the boundary or the separation of points of blockage on grain boundaries. In practice this condition is not satisfied in superplastic cavitation by virtue of both the small values of L and the low applied stresses. An important factor here may then be the segregation of trace impurities to grain boundaries [9].

The role of grain boundary particles in cavity nucleation is better understood. Grain boundary sliding is necessary to induce the initial breakdown of the particle/matrix interface, and this is reflected in the observation that cavitation is a maximum when strain rate sensitivity is highest, indicating a maximum in grain boundary sliding [5]. A model for cavitation at particles has been presented by Smith and Barnby [16], but the model relies on fracture of the particles, which is not observed in practice, and also predicts that small particles are preferential sites, whereas the opposite is found in practice. A model embodying the Smith-Barnby and the Stroh criteria based on decohesion of the particle has been developed by Fleck, Beevers and Taplin [17]. This model predicts nucleation preferentially at large particles and is generally consistent with experimental observations. Thermodynamic considerations indicate that there is a minimum radius at which a cavity nucleus is stable against shrinkage, and this implies that there is a minimum effective particle size. In superplastic deformation grain boundary sliding is likely to be the dominant factor in cavity stabilization, whereas vacancy absorption is probably more important in creep.

The growth of voids greater than the initial size may then be controlled by diffusion or deformation, and, at the superplastic strain rates used here, deformation seems to be the dominant factor. This is consistent with the irregular shapes of the cavities, which grow rapidly to become single facet cracks. Hancock [18] has studied the conditions under which hole growth without vacancy condensation should be faster than growth by diffusion. He shows that diffusional hole growth is favoured when holes are small and the value of $\sigma/\dot{\varepsilon}$ is high. For example, growth by diffusion in a ferritic steel should be the slower mechanism for hole radii > 0.7 µm when $\sigma/\dot{\varepsilon}$ = 9 MNs mm^{-2}. Typical values of $\sigma/\dot{\varepsilon}$ used in the work reported here range from 6.8 x 10^{-2} to 2.8 x 10^{-4} MNs mm^{-2} so that void growth should be deformation controlled almost throughout, assuming that there are no order of magnitude differences in the relevant material parameters between ferritic steel and Al-Mg. It was also shown that the deformation controlled

growth of holes at high temperature follows McClintock's [19] plasticity solutions for void growth in low temperature ductile fracture, emphasizing the essential similarity in the processes.

The main distinguishing feature of superplastic deformation in a cavitating material is then its resistance to cavity linkage by internal necking, which is allied to the resistance to the formation of small radius external necks which is common to cavitating and non-cavitating alloys alike. A complete spectrum of macroscopic necking behaviour is now recognized, ranging from negative strain rate susceptibility [20], where the formation of intense shear bands gives failure at very low strains, through conventional low temperature ductile behaviour with no strain rate effects to the positive rate susceptibility of superplasticity. The problem of stability against external necking has been addressed by Hutchinson and Neale [21]. They demonstrate the inherent weakness of linearized analyses such as those of Hart [22], which do not satisfactorily predict the tensile behaviour of materials with a strong strain rate dependence e.g., the results of Sagat and Taplin [23], as necking is essentially non-linear. They present a non-linear analysis, which they describe as a long wavelength approximation. On the basis of their analysis large uniform elongations are shown to be clearly predictable in strain rate sensitive materials, and even low values of m ($\sim$ 0.1) should substantially enhance values of fracture strain. This should provide added impetus to investigate the potential for superplastic deformation of other alloys of plebeian stock.

Resistance to failure by internal necking between cavities *per se* has not been studied as closely. Two potential failure mechanisms may be distinguished. Sufficient cavities may coalesce to constitute a crack, which then propagates by tearing. This may be likened to the conventional model of low temperature ductile fracture by internal necking followed by shear. Alternatively, the reduction in effective cross section by cavity growth may be such that failure effectively occurs simultaneously over the whole section, which is categorized as void sheeting. Metallographic observation of the material used in the present work appears marginally to suggest the operation of the first mechanism, see Figure 2. Void interlinkage has also been observed in aluminum bronze and tertiary brass [2, 7]. In Zn-22% Al internal linkage of voids by the intergranular void sheet process has been reported [14]. In tests at a range of temperature on an aluminum bronze a transition in fracture mechanism with temperature from ductile rupture to intergranular void sheet and finally to propagation and interlinkage of grain boundary cracks was found [3]. This result illustrates the important effect that increasing temperature may have on the cavitation process, as a consequence of the reduction in grain stability, where boundary migration may relax stress concentrations and reduce cavity nucleation.

Models of ductile fracture by the growth and coalescence of voids have been developed by McClintock [19], Brown and Embury [24] and Thomason [25]. They concentrate essentially on void growth and predict instability when a critical void spacing is achieved, whereas it is felt that the process of coalescence and final failure is more crucial in the present study. Thomason [26] has extended his analysis to include effects of strain rate sensitivity in order to model both blue brittleness and superplasticity in a cavitating material. As deformation proceeds the strain rate in the bulk material for uniform flow should be significantly lower than that in the ligaments between cavities. Hence void coalescence should be inhibited in materials having marked strain rate sensitivity with a consequent increase in total elongation as deformation is transferred from the ligaments to the necks. For a given volume fraction of cavity nucleating sites a greater elongation to fracture should be achieved with finer dispersions as this produces more inter-cavity ligaments with higher strain rates.

The general model of the fracture process is then as indicated in Figure 3. Cavities nucleate at particles in grain or phase boundaries by boundary sliding. There may be some initial growth by vacancy condensation, but the dominant growth mechanism as cavities grow to become single facet cracks rapidly becomes plastic strain accumulation. The main distinguishing feature of materials with significant strain rate sensitivity is then found in the resistance to necking and failure of the inter-cavity ligaments. An important point not yet resolved is the mechanistic reason for strain rate sensitivity in the internal necks of a cavitating superplastic material. In the bulk this sensitivity arises from deformation by grain boundary sliding, but the inter-cavity material can probably at best be considered only a bi-crystal, and some further explanation is necessary, possibly in terms of the increased flow stress for non-superplastic deformation.

It is of great importance to note that failure by cavitation can often take place very rapidly, in some cases in a matter of minutes. This is counter to the view of Cottrell [27] regarding the Flixborough failure. Results such as those discussed here should be seen in the context of cavitation as a generic fracture mechanism, not as a process specific to long term creep behaviour, but significant in welding, hot working, superplastic forming, thermal cycling as well as creep and fatigue at high temperatures. In these circumstances rapid processes of creep cavitation are more readily understood.

CONCLUSIONS

(1) Superplastic elongations greater than 400% may be achieved in a range of Al-Mg alloys by simply controlling grain sizes and selecting critical strain rates and deformation temperatures.

(2) The alloys fail by growth and interlinkage of cavities without the formation of an external crack.

(3) A qualitative model of the fracture process is derived on the basis of models for high and low temperature fracture. The mechanism by which cavity linkage is suppressed in a strain rate sensitive material remains to be resolved.

(4) Short time failure by cavitation is not uncommon and may be of great practical significance.

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Table 1 Alloy Composition (wt. %)

Alloy	Mg	Mn	Cr	Bi	Fe	Si
A	4.98	0.94	0.10	-	0.27	0.16
B	4.95	0.49	0.10	-	0.27	0.16
C	4.47	0.73	0.074	0.006	0.32	0.17
5083	4.5	0.70	0.15	-	<0.4	<0.4

Table 2 Tensile Test Results: Effect of Temperature, $\dot{\varepsilon} = 2.6 \times 10^{-4}\ s^{-1}$

Alloy	Test Temperature, K	% Elongation to Fracture	Approx. Optimum Temp., K
A	573	142	813
	673	145	
	773	231	
	813	462	
	823	363	
	833	222	
B	573	167	823
	673	209	
	773	350	
	798	315	
	823	376	
	843	145	
C	573	190	773
	673	323	
	773	387	
	823	283	

Table 3 Tensile Test Results: Effect of Strain Rate at Optimum Temperature

Alloy	Test Temperature K	$\dot{\varepsilon}$ s^{-1}	% Elongation to Fracture	Approximate Optimum $\dot{\varepsilon}$, s^{-1}
A	813	2.6×10^{-3}	303	2.6×10^{-4}
		2.6×10^{-4}	462	
		6.4×10^{-5}	307	
B	823	2.6×10^{-3}	114	1.2×10^{-4}
		2.6×10^{-4}	376	
		6.4×10^{-5}	381	
C	773	6.4×10^{-3}	294	3.0×10^{-4}
		2.6×10^{-4}	387	
		6.4×10^{-5}	264	

Table 4 Area Fraction of Cavities: Effects of Temperature and Strain Rate

Alloy	Temperature K	Strain Rate s^{-1}	% Elongation to Fracture	Area Fraction of Cavities
A	573	2.6×10^{-4}	142	8.1
	773		231	14.6
	813		462	25.4
	833		222	13.1
	813	2.6×10^{-3}	303	13.1
		2.6×10^{-4}	462	25.4
		6.4×10^{-5}	307	15.2
B	798	2.6×10^{-4}	315	10.4
	823		376	20.6
	843		145	14.5
	823	2.6×10^{-3}	114	3.5
		2.6×10^{-4}	376	20.6
		6.4×10^{-5}	381	17.0

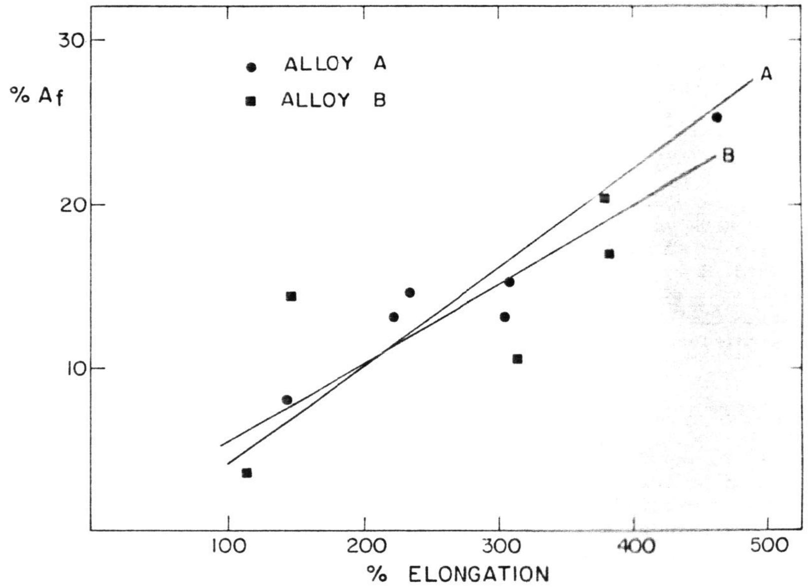
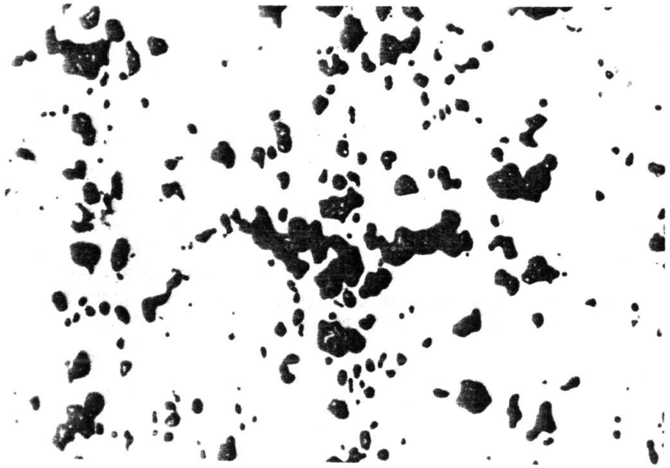


Figure 1 Area Fraction of Cavities, A_f, versus Elongation for Alloys A and B

(a) X 16

(b) X 16



(c) X 70

Figure 2 The Final Stage of Cavity Coalescence

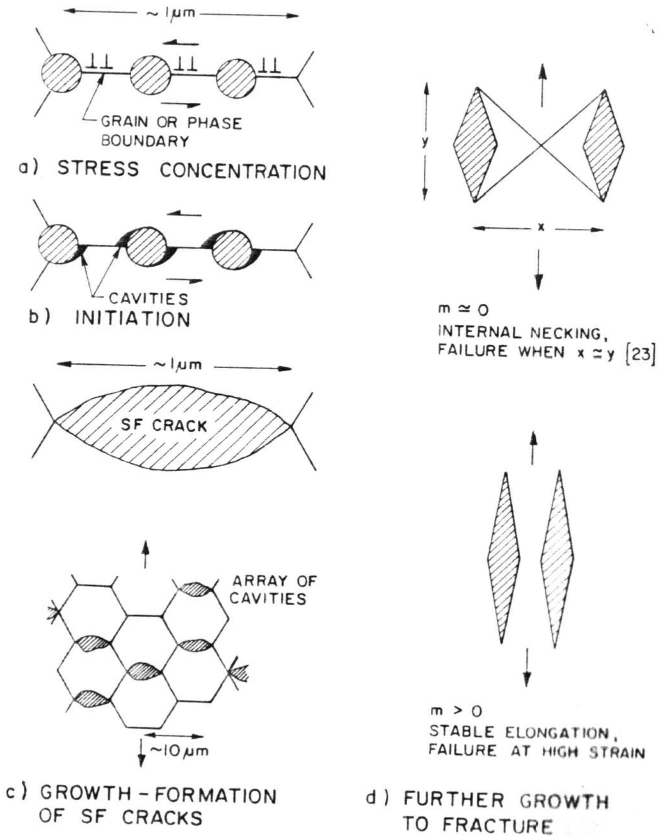


Figure 3 Model of the Cavitation and Failure Process