INFLUENCE OF MICROSTRUCTURE ON LIQUID METAL EMBRITTLEMENT OF ALUMINUM ALLOYS

R. G. Hoagland,* Y. Liu,* and B. Benson*

This paper summarizes several experimental observations of cracking behavior and fracture topography that reveal certain aspects of the influence of microstructure of aluminum and its alloys when embrittled by liquid mercury and gallium. Two properties of the solid which are particularly important to the embrittlement are grain shape and orientation relative to the crack plane, and the strength level. These results are interpreted from the standpoint of screening, loss of constraint, and differences in crack driving forces needed to sustain intergranular fracture relative to cleavage.

INTRODUCTION

Liquid gallium and mercury can cause a very substantial decrease in the load carrying capability of aluminum and its alloys particularly when the liquid has access to the tip of a crack or similar type of flaw in the solid. Indeed, the effect of these two liquids on aluminum is one of the most dramatic examples of embrittlement by environmentally assisted slow crack growth, in which the load carrying ability is reduced as much as a hundredfold. The mechanism of embrittlement involves the migration of liquid metal atoms to the crack tip where they promote separation of the solid when the solid is subject to a tensile stress tending to open the crack. Beyond this general guideline, various explanations of the interactions in this and similar forms of embrittlement in other liquid/metal systems differ in the literature, as shown for example, by the extensive review of this topic by Stoloff (1) which describes mechanisms that include enhanced decohesion, enhanced shear, dissolution, and intergranular diffusion. Common agreement concerning the nature of the embrittling mechanism has not yet evolved, but the enhanced decohesion model is attractive on physical grounds. In general, decohesion models are concerned with a reduction in the interatomic forces between the atoms at a crack tip thereby enabling crack extension at a lower tip J than occurs in air (or vacuum). However, the issue as to which model is most appropriate is clouded by the fact that microstructural properties of the solid and the composition of both the solid and liquid affect cracking response, and such factors vary among the many experiments which have been reported.

This paper presents a description of several microstructural factors that influence the cracking behavior of aluminum alloys in contact with mercury and gallium. Our results focus primarily on effects related to crack path and strength of the solid and derive from experiments measuring the stress intensity dependence of the crack velocity and from observations of the fracture appearance. We suggest that the driving force for a moving crack, characterized by a local or tip J, may be substantially different from the applied J, since the crack tip can be screened from

* Dept. of Mech. and Matls. Eng'g, Washington State Univ., Pullman, WA 99164

the applied J in several ways.

EXPERIMENTAL METHODS

The results presented here derive primarily from experiments on a high strength aluminum alloy, 7075, which contains about 6 wt% Zn, 2.5 wt% Mg, 1.5 wt% Cu, 0.3 wt% Mn, and lesser amounts of Fe, Cr, Si, and Ti. This material was obtained in the form of 12.5 mm thick plate in the peak aged T651 condition, and possessed an elongated grain structure with average dimensions in the rolling, long transverse, and short transverse directions of 300 μm, 100 μm, and 20 μm, respectively. Yield and tensile strength of the as-received plate are about 511 MPa and 572 MPa, respectively, and K_{Ic} in air is about 25 MPa√m and 18 MPa√m in the T-L and S-L orientations, respectively. Lower strength levels were prepared by overaging. Some experiments were also conducted on 99.999% pure, large grain (2mm) Al.

Crack growth experiments were performed on double cantilever beam geometries in which load and load-point displacement were continuously recorded to provide time correlated K, crack length, a, and therefore, crack velocity, da/dt. Specimens taken from the rolled plate such that crack propagation was in the rolling direction, and crack plane perpendicular to the short transverse direction (the T-L orientation) had dimensions sufficient to achieve K_{Ic} while maintaining small scale yielding. Some tests were also conducted on specimens with an S-L orientation (crack plane perpendicular to the short transverse direction) and these specimens were smaller and small scale yielding requirements limited K in these specimens to levels below about 15 MPa√m. The test was begun within a few seconds after the liquid metal was supplied to the premachined slot tip from a small reservoir in the specimen. Further experimental details can be found in Refs 2 and 3.

RESULTS

We present results showing the influence of flow strength and crack plane orientation relative to the textured grain structure, and, we also summarize the fracture appearance which is described in somewhat greater detail in Refs 2 and 3.

Strength and Orientation Effects

The da/dt - K results for 7075 in Figure 1a show examples of the influence of strength on the response of the T-L orientation in Hg, and in Figure 1b compare the responses of the T-L orientation with the S-L orientation in Ga for the peak aged material. The results in Figure 1a were obtained with fixed load, i.e. K increased as the crack grew, while in Figure 1b, the data are primarily from fixed displacement tests with K decreasing with increasing crack length. The results display the typical Γ-shaped curve characterized by a plateau of approximately K independent crack growth and a very strongly K-dependent segment that appears as a threshold. From repeated tests on individual specimens and on groups of specimens, it has been shown (2-4) that such results are not very repeatable, particularly with regard to the threshold which may vary by as much as ±30%. Other parts of the Γ-curve also display test-to-test variability, and, in addition, significant hysteresis results between segments of the curve on changing from increasing K (fixed load) to decreasing K (fixed displacement). However, the trends represented by the data in

Figure 1 are accurate, with [1] a shift in the Γ-curves to higher K-levels as strength decreases, [2] much lower thresholds for the S-L orientation, and [3] lower thresholds (more severe embrittlement) when the crack is in contact with gallium than with mercury.

The trends suggested by Figures 1a and 1b are summarized in Figure 2 which shows the K needed to achieve a particular crack velocity (in this case 0.001 m/s) as a function of yield strength for both the T-L and S-L orientations of 7075 with Hg present. In both orientations, increasing strength reduces the K needed to sustain this crack velocity. Also, the difference in response between the T-L and S-L orientations is quite extraordinary, and the origin of these differences is related to the crack path and fracture mode, which, as noted below, is mixed cleavage and intergranular involving ductile ligaments for the T-L orientation, but entirely intergranular with no ligaments for the S-L orientation.

Fracture Mode

When aluminum and nearly all of its alloys are subjected to tensile fracture in air or other similarly benign environment, the fracture proceeds by a void nucleation and growth mechanism producing a ductile-dimple fracture surface that is typical of the fracture of tough FCC metals. When embrittled by gallium or mercury, however, the fracture mode undergoes a transition to other modes depending upon the orientation of the crack plane relative to the microstructural texture (2,3). In the T-L orientation wherein the crack plane and crack front are, on average, parallel to the shortest dimension of the grains in the 7075, the crack grows predominantly by cleavage, although intergranular delaminations perpendicular to the mean crack plane are common, and occasional patches of elongated ductile dimples are also observed (2). Two, and even three, cleavage planes are occasionally found to intersect within a single grain (3), and, as they are always seen to be orthogonal, it appears that the cleavage plane is probably {100}. No fine-scale dimples are observed on the cleavage planes. The patches of elongated dimples are related to ligaments, or islands of temporarily unbroken material, left behind by the growing crack (2) which eventually fail by a ductile mode. The occurrence of these ligaments is stochastic and probably accounts for the test-to-test variability in the da/dt - K results.

In the S-L orientation, the crack plane is favorably oriented for propagation on the grain boundaries of the flattened grains in 7075, and, indeed, the fracture is observed to be almost entirely intergranular in both Hg and Ga. Moreover, fracture of large-grain, pure aluminum proceeds completely intergranularly and no evidence of ligaments is found. Thus, relative to cleavage, the grain boundary consumes less energy during separation (and, therefore, a requires a lower K) than the cleavage plane, in the presence of these liquid metals. In fact, given enough time, liquid gallium is observed to penetrate the grain boundaries of aluminum at zero stress, which suggests (5,6) that the interfacial energy between Ga and Al is less than one half the grain boundary energy of Al.

DISCUSSION

We propose that, depending only on fracture mode, the energy dissipation rate at the crack tip, which is the crack tip driving force, J_{tip}, is essentially constant while the crack is advancing. In this view then, the effect of microstructure and strength becomes attributable to the difference in J_{tip} for intergranular and cleavage fracture, to screening effects related to the presence of sources of internal stress (other than the crack tip), and to effects related to loss of constraint. The following describes these factors in additional detail.

Plastic Wake Effects

The plastic wake is the plastically deformed material left behind the extending crack, and contains residual stresses which, in turn, alter the crack driving force, i.e. the effect of a wake is to screen the crack. The size (thickness) of the wake increases with decreasing strength, and, therefore, it is natural to expect that the difference in the apparent driving force, $J_{applied}$, which is equal to K^2/E, and the J_{tip} depends on yield strength. This is an example of a situation wherein the J-integral is no longer path independent. The ratio $J_{applied}/J_{tip}$ becomes larger than unity and should increase with increasing plastic zone size (7,8). In order to gain a more quantitative sense of the dependence of wake screening on yield strength, we have modeled the wake as an array of superdislocations, extending back from the crack tip. This array exerts a force on the crack tip from which a relation between $J_{applied}/J_{tip}$ and the yield strength can be estimated (9). The further assumption, that J_{tip} is a constant and, therefore, independent of strength leads to a prediction for the strength dependence of the data shown in Figure 2, and identified as a dashed line. The magnitude of the wake effect according to this model is not strong, increasing the K required to maintain crack propagation by about 25% for a sixfold decrease in strength from 600 MPa to 100 MPa, but the agreement with the S-L results is reasonably good. When applied to the T-L case, the model predicts a weaker dependence on strength than is observed. As we have noted above, however, cracking behavior in the T-L orientation is somewhat more complex than the S-L.

Delamination

Delaminations caused by penetration of liquid metal along grain boundaries approximately perpendicular to the average crack surface are commonly observed on the fracture surfaces of the T-L oriented specimens but not on the S-L fracture surfaces. These delaminations are a consequence of plane strain, but promote a transition to plane stress because the material becomes, in effect, a multilayer composite of thin sheets. This transition from plane strain to plane stress increases the plastic zone size, an effect which is enhanced by decreasing strength.

Ligaments

An unbroken ligament supplies pinching forces that also screen the crack tip. The magnitude of this screening may be quite strong particularly if the ligament applies increasing forces to the crack faces as it stretches. Ligament formation and rupture is observed in the fracture of T-L orientations but not the S-L, nor was it observed in the fracture of pure aluminum. The mechanism that stabilizes a patch of material, at least temporarily, in the embrittling liquid and allows the crack to move leaving it behind, may be related to oxidation (4) at the tip and to the formation of shear walls (2). If ligament failure occurs by a stress controlled mechanism, as has been suggested (10) in cleavage/ligament rupture in bcc ferrous materials, then one expects that ligament survival is improved with decreasing flow strength, and as a result, ligaments would become more effective in screening the crack tip. The question of ligament failure mechanism needs to be established. Nevertheless, it seems likely that ligaments probably account for the much higher K required for cracking in the T-L orientation than the S-L. In addition, the creation and rupture of ligaments are stochastic events which may also account for the lack of repeatability of the da/dt - K results.

CONCLUSIONS

We have presented some results that focus attention primarily on the role played by microstructure in the cracking of an aluminum alloy in contact with Ga and Hg. Flow strength and crack orientation relative to grain morphology are seen to exert a substantial influence on the crack growth behavior. We suggest that the effects derive from: [1] screening from unbroken ligaments and plastic wake, the latter being a somewhat weaker effect, [2] loss of constraint because of delamination due to liquid penetration along grain boundaries oriented approximately perpendicular to the average crack plane, and [3] the difference in energy dissipation rates for cleavage compared to intergranular separation.

ACKNOWLEDGEMENTS

This work was supported by the Division of Materials Science, Office of Basic Energy Sciences, United States Department of Energy under Grant DE-FG06-87ER45287.

REFERENCES

1. Stoloff, N. S., in "Atomistics of Fracture", Edited by R. M. Latanison and J. R. Pickens, NATO Conf. Proc., 1982, pp. 921-949.

2. Liu, Y. and Hoagland, R. G., Scripta Met., Vol. 23, 1989, pp. 339-344.

3. Benson, B. A. and Hoagland, R. G., Scripta Met., Vol. 23, 1989, pp. 1943-1948.

4. Wheeler, D. A., Hoagland, R. G., and Hirth, J. P., Corrosion, Vol. 45, 1989, pp. 207-212.

5. Goddard, D. J. and Williams, J. A., Journ. Inst. Metals, Vol. 99, 1971, p. 323.

6. Kargol, J. A. and Albright, D. L., Met. Trans., Vol. 8A, 1977, p. 27.

7. Thompson, R. and Sinclair, J. E., Acta Metall., Vol. 30, 1982, p. 1325.

8. Hirth, J. P., Hoagland, R. G., and Popelar, C. H., Acta Metall., Vol. 32, 1984, p. 371.

9. Liu, Y. and Hoagland, R. G., unpublished research.

10. Hoagland, R. G., Hahn, G. T. and Rosenfield, A. R., Metall. Trans., Vol. 3, 1972, p. 123.

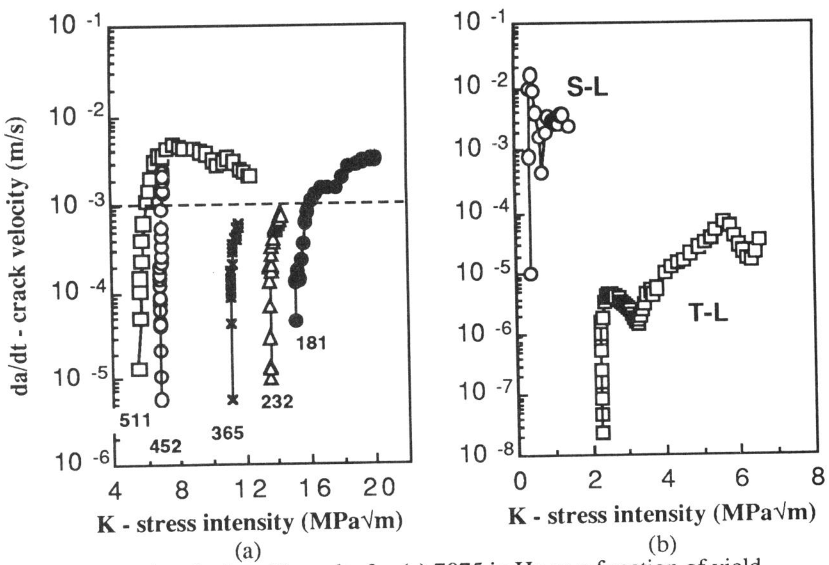

Figure 1. Crack velocity - K results for (a) 7075 in Hg as a function of yield strength levels (given in units of MPa) from increasing K tests, and (b) for peak aged 7075 (yield strength = 511 MPa) in Ga for two different cracking orientations.

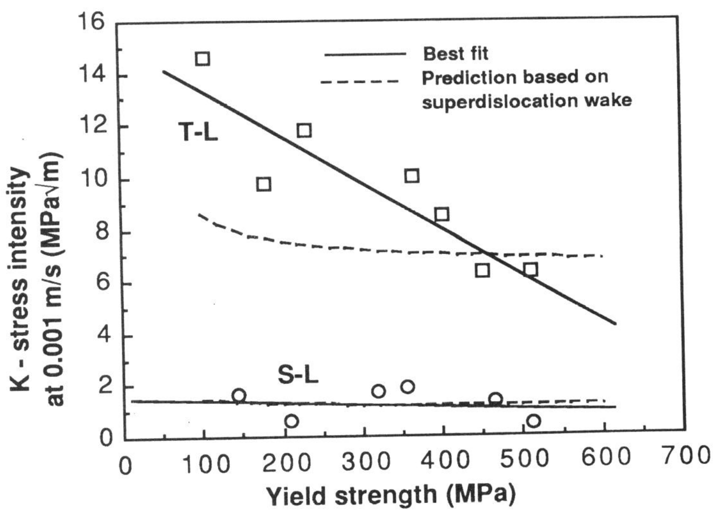

Figure 2. Effect of yield strength on the K needed to sustain a da/dt of 0.001 m/s in Hg. This graph shows experimental results for two specimen orientations and a prediction based on a superdislocation model of the plastic wake.