THE FRACTURE CHARACTERISTICS OF Al AND Al-5% Mg AT ELEVATED TEMPERATURES

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ABSTRACT

Experiments were conducted on high purity Al and an Al-5% Mg solid solution alloy to determine the fracture behavior at elevated temperatures. Two types of test were carried out. First, specimens were pulled to fracture over a range of nominally constant strain rates at a temperature of 573 K. Second, specimens were pulled to fracture over a wide range of temperatures at a strain rate of 3.3×10^{-4} s^{-1}. The results show that there are very important differences in the fracture characteristics of pure Al and Al-5% Mg. Pure Al does not exhibit intergranular fracture, at least over the range of stresses and strain rates investigated experimentally, and the specimens fail primarily in a transgranular manner at 573 K with a fracture strain, ε_f, which is relatively independent of the testing strain rate. By contrast, the addition of 5% Mg in solid solution leads to two maxima in ε_f at 573 K, at $\sim 10^{-5}$ and $\sim 3 \times 10^{-3}$ s^{-1}, respectively. These maxima correspond to transgranular fracture, and there is also a ductility minimum at $\sim 3 \times 10^{-4}$ s^{-1} due to intergranular fracture. These transitions give rise to marked differences in the appearance of the Al-5% Mg specimens after fracture. It is shown that the results on high purity Al are in good agreement with an experimental fracture map.

KEYWORDS

Transgranular fracture; intergranular fracture; ductile fracture; rupture; aluminum; Al-Mg alloys; fracture maps.

INTRODUCTION

Polycrystalline metals may fracture by several different and distinct mechanisms, and the fracture mode depends both on the testing conditions and on various material parameters such as the grain size and the concentration of precipitates. There are three basic types of fracture under creep conditions at elevated temperatures.

First, a polycrystalline material may break by a *transgranular* process, due to the nucleation and growth of holes at inclusions in the matrix and the subsequent occurrence of internal necking and hole coalescence. This process is qualitatively similar to the mechanism of *ductile fracture* observed at low temperatures (McClintock, 1968).

Second, a polycrystal may fail in an *intergranular* manner by separation along the grain boundaries. Intergranular creep fracture occurs either by the formation and growth of wedge cracks at the triple points or by the nucleation, growth, and interlinkage of grain boundary cavities. Several possible processes of cavity growth have been considered theoretically, and a unified approach has been developed to incorporate these various processes (Miller and Langdon, 1980).

Third, a polycrystalline specimen may neck down to a point or chisel edge at high temperatures and rapid strain rates, to give high temperature *rupture*.

Finally, it should be noted that there is also a fourth process, termed *dynamic fracture*, which occurs under dynamic loading conditions at strain rates greater than $\sim 10^6$ s^{-1}: this process is generally of little interest in creep experiments.

This paper describes the results obtained in a series of experiments designed to investigate and compare the fracture characteristics of a high purity metal and a dilute solid solution alloy. The materials selected for this work were Al and Al-5% Mg. These two metals were chosen because it is well established that aluminum tends to fail transgranularly or by rupture (Chin *et al.*, 1964), and there are no experimental reports of intergranular failure in high purity Al under high temperature creep conditions (Grant, 1959). However, there is a transition from transgranular to intergranular fracture in Al of low purity, typically $\sim$99% Al (Servi and Grant, 1951), and there is also very extensive evidence for intergranular failure in Al-Mg solid solution alloys (Mullendore and Grant, 1954). For example, Al alloys containing, respectively, 1%, 2%, and 5% Mg were reported to fail intergranularly at a testing temperature of 533 K, with the extent of intercrystalline cracking becoming increasingly extensive as the Mg content was raised. For the Al-5% Mg alloy, it was reported that intergranular failure occurred over a wide range of strain rates at 533 K, but intergranular cracking was restricted to fast strain rates at 644 K and was not observed over the creep rates examined experimentally at 755 K. An interesting result reported by Mullendore and Grant (1954), for all three Al-Mg alloys tested at 533 K, was the presence of a pronounced minimum in a plot of the elongation up to the end of the secondary stage of creep versus the total rupture time. This minimum was tentatively attributed to restricted grain boundary migration; however, the overall influence of strain rate on the strain to fracture, ε_f, has not been examined in any detail.

As will be demonstrated in the present investigation, there is a marked variation in ε_f with strain rate for Al-5% Mg tested at 573 K under conditions of nominal constant strain rate, and this variation is associated with very significant changes in the fracture mode. Furthermore, these changes may be revealed very easily by visual examination of the specimens after fracture.

EXPERIMENTAL MATERIALS AND PROCEDURE

Two different materials were used in this work: a high purity (99.995%) aluminum and an Al-5.25 wt % Mg solid solution alloy (equivalent to Al-5.8 atomic % Mg and herein designated Al-5% Mg). Both materials were supplied in the form of rods, 12.7 mm in diameter, and tensile specimens were cut from the rods with a reduced gauge length, circular in cross-section, of 25.4 mm.

The high purity Al specimens were annealed for 24 hours at 673 K to give an average initial grain size, d, of about 300 μm determined by the linear intercept method. The Al-5% Mg specimens were annealed for 1.75 hours at 813 K to give an average grain size of about 575 μm.

All tests were conducted in air on an Instron machine operating at a constant rate of cross-head displacement: all quoted strain rates refer to the initial rates at zero strain. A three-zone split furnace was used to obtain the required testing temperature, and the temperatures were held constant during each test to within ±1 K of the reported values. Two types of test were conducted: first, specimens were pulled to fracture over a range of nominally constant strain rates at a constant temperature of 573 K, corresponding to ~ 0.6 T_m where T_m is the absolute melting point; second, specimens were pulled to fracture over a wide range of temperatures at a constant initial strain rate of 3.3×10^{-4} s^{-1}.

EXPERIMENTAL RESULTS

Figure 1 shows the fracture results obtained on high purity Al tested at 573 K, plotted as the total strain at fracture, ε_f, versus the initial strain rate, $\dot{\varepsilon}_i$. The plot shows that ε_f is essentially independent of strain rate in the range of $\sim 10^{-6}$ to 10^{-3} s^{-1}, and there is only a very slight decrease in ε_f at $\dot{\varepsilon}_i > 10^{-3}$ s^{-1}: the latter decrease may be due to the exceptionally high stresses occurring at these rapid rates ($\sigma/G > 2 \times 10^{-3}$, where σ is the applied stress and G is the shear modulus). This result is qualitatively in agreement with the observation of Servi and Grant (1951), on an aluminum of similar purity, that the elongation to the end of the secondary stage of creep is independent of the total rupture time at 573 and 644 K.

All of the high purity Al specimens tended to fail transgranularly at 573 K within the testing range of $\dot{\varepsilon}_i = 3.3 \times 10^{-6}$ to 3.3×10^{-2} s^{-1}, although there was a trend towards high temperature rupture at the fastest rates. An example of a fractured specimen is shown in Fig. 2 for a test conducted at $\dot{\varepsilon}_i = 3.3 \times 10^{-4}$ s^{-1} to $\varepsilon_f = 1.4$. It is evident from the fracture tip of this specimen that necking occurred and the cross-sectional area reduced almost to a point before final failure.

A similar set of results is shown in Fig. 3 for tests conducted on Al-5% Mg over the same range of initial strain rates. The behavior in this case is very

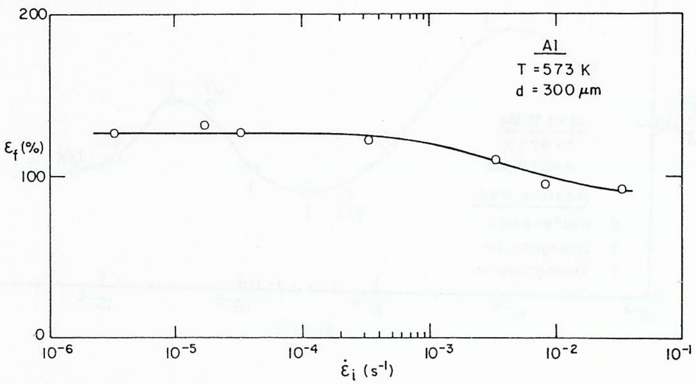

Fig. 1. Strain at fracture, ε_f, versus initial strain rate, $\dot{\varepsilon}_i$, for high purity Al at 573 K.

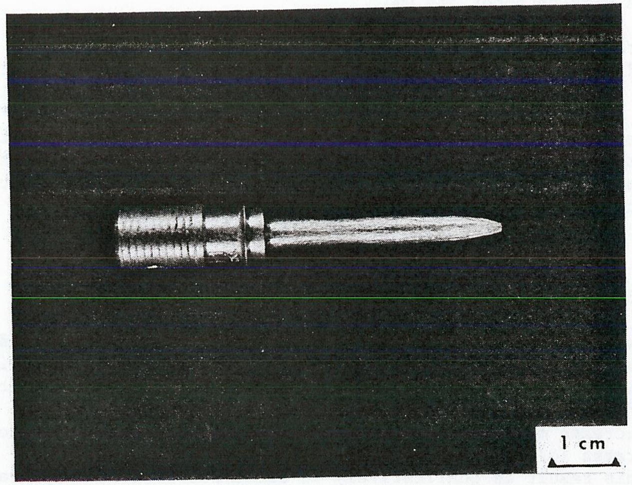

Fig. 2. Fracture tip of high purity Al specimen tested at $\dot{\varepsilon}_i = 3.3 \times 10^{-4}\ s^{-1}$ and T = 573 K to fracture at $\varepsilon_f = 1.4$.

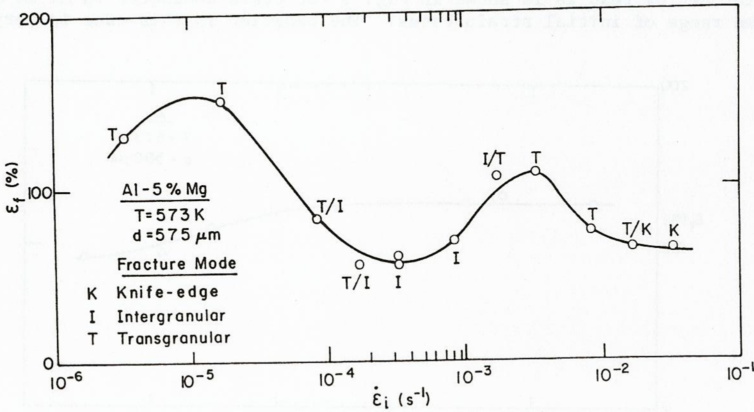

Fig. 3. Strain at fracture, ε_f, versus initial strain rate, $\dot{\varepsilon}_i$, for Al-5% Mg at 573 K. The letters designate the fracture modes.

different, and there are two clear maxima in ε_f at $\sim 10^{-5}$ and $\sim 3 \times 10^{-3}\ s^{-1}$, respectively. The various letters in Fig. 3 designate the fracture mode determined by examination of the specimens after failure. It is clear from these letters that the two peaks in ε_f correspond to conditions where fracture occurred transgranularly (T), whereas the ductility minimum at $\dot{\varepsilon}_i \simeq 3 \times 10^{-4}\ s^{-1}$ is associated with intergranular fracture (I) and the drop in ductility at very rapid strain rates represents a transition to a chisel or knife edge type of failure (K). Experimental specimens marked by two letters (T/I, T/K, etc.) exhibited fracture behavior which appeared to be a mixture of the two designated modes.

The marked changes in behavior at the different strain rates may be very readily appreciated by a comparison of the fractured specimens of Al-5% Mg, as illustrated in Fig. 4. The specimen shown in Fig. 4(a) failed in a transgranular manner after testing at $\dot{\varepsilon}_i = 1.67 \times 10^{-5}\ s^{-1}$ to $\varepsilon_f = 1.5$. By contrast, Fig. 4(b) shows a specimen tested to $\varepsilon_f = 0.60$ at $\dot{\varepsilon}_i = 3.3 \times 10^{-4}\ s^{-1}$. This specimen failed intergranularly with little or no necking at the point of fracture, so that the final fracture surface has a substantial cross-sectional area. This striking result was confirmed with a duplicate specimen, as indicated by the two experimental points shown for $\dot{\varepsilon}_i = 3.3 \times 10^{-4}\ s^{-1}$ in Fig. 3. Finally, Fig. 4(c) shows the specimen tested to $\varepsilon_f = 1.06$ at $\dot{\varepsilon}_i = 3.3 \times 10^{-3}\ s^{-1}$. This specimen again failed transgranularly, with necking and final failure almost at a point. A comparison of the specimens shown in Figs. 4(a) and (c) confirms the similarity in behavior at these two different strain rates.

After fracture, several of the Al-5% Mg specimens were sectioned longitudinally and prepared for metallographic examination. Some internal cavitation was evident at all strain rates.

Finally, several high purity Al specimens were pulled to fracture at $\dot{\varepsilon}_i = 3.3 \times 10^{-4}\ s^{-1}$ and at different temperatures up to 773 K. An examination of the fractured specimens showed a transition from an essentially ductile fracture at room temperature (298 K) to transgranular failure at the higher temperatures.

DISCUSSION

The results obtained in this investigation confirm that there are very important differences in the fracture characteristics of high purity Al and the Al-5% Mg alloy. Intergranular fracture is not observed in high purity Al, at least under the testing conditions used in this work, and the dominant failure mode is transgranular. As a result, there is very little variation in fracture strain with the testing strain rate, and a series of specimens tested at 573 K exhibited fracture strains of ~90-125% over four orders of magnitude of initial strain rate. By contrast, the behavior of Al-5% Mg is more complex at 573 K, with two maxima in fracture strain associated with transgranular fracture and an intermediate minima due to the occurrence of intergranular failure. This behavior leads to a very marked change with strain rate in the appearance of the specimens after fracture, as documented in Fig. 4.

Mullendore and Grant (1954) reported, from their experiments on various Al-Mg alloys, that the elongation to the end of the secondary stage of creep decreased at constant rupture time with increasing Mg content. However, the present results, conducted under nominal constant strain rate rather than creep (constant stress) conditions, show that the fracture strains in the transgranular regions in Al-5% Mg tend to be at least as high or higher than in high purity Al. For example, the fracture strains at $\dot{\varepsilon}_i \simeq 10^{-5}\ s^{-1}$ are ~150% (Fig. 3) and ~125% (Fig. 1) in Al-5% Mg and high purity Al, respectively.

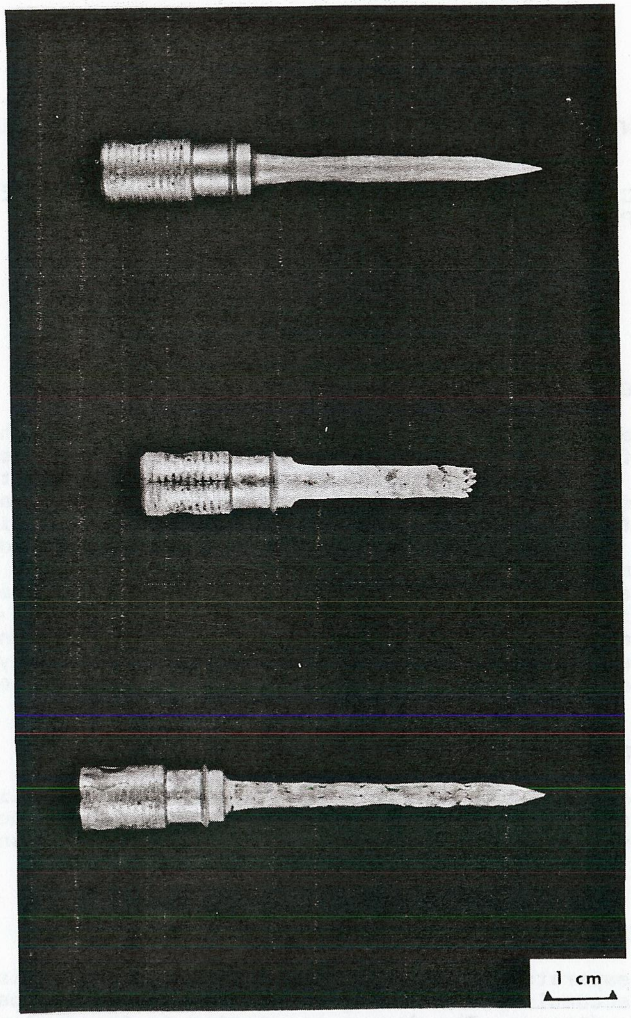

Fig. 4. Fracture tips of Al-5% Mg specimens tested at T = 573 K under the following conditions:
(a) $\dot{\varepsilon}_i = 1.67 \times 10^{-5}\ s^{-1}$ and $\varepsilon_f = 1.5$;
(b) $\dot{\varepsilon}_i = 3.3 \times 10^{-4}\ s^{-1}$ and $\varepsilon_f = 0.60$;
(c) $\dot{\varepsilon}_i = 3.3 \times 10^{-3}\ s^{-1}$ and $\varepsilon_f = 1.06$.

The various fracture mechanisms occurring in any material may be depicted graphically by plotting a fracture map. There are two basic types of fracture map (Miller and Langdon, 1979): an *experimental map* in which a large number of experimental points from several different investigations is used to identify the fracture modes in stress-temperature space (Ashby, Gandhi and Taplin, 1979), and a *model-based map* in which the various times to fracture are numerically evaluated for each of the possible fracture processes (Ashby, 1978). The latter type of map is clearly advantageous, although satisfactory theoretical analyses may not be available for all of the fracture mechanisms (for example, high temperature rupture). Several examples of the model-based approach are now available (Ashby, 1978; Langdon and Miller, 1979; Miller and Langdon, 1979).

Figure 5 shows the experimental fracture map derived by Ashby, Gandhi and Taplin (1979) for high purity (99.9%) aluminum, plotted as σ/E versus the homologous temperature, T/T_m, where E is Young's modulus and T is the testing temperature. This map was obtained for Al having a normal range of grain sizes, and should be reasonably applicable when d = 300 μm. The solid lines on the map separate the three fields of ductile fracture, transgranular fracture, and rupture: the broken line indicates the approximate transition to conditions of dynamic rupture. Superimposed on the map are the experimental points obtained in the present investigation, with each fracture type separately identified.

It is clear that the present results are in very good agreement with the fracture map for high purity Al, especially when it is noted that there are regimes of mixed modes along each of the field boundaries. It is also clear that the data for Al-5% Mg (Fig. 3) would not fit on the map because of the advent of intergranular failure at intermediate strain rates. This point is important because an examination of the

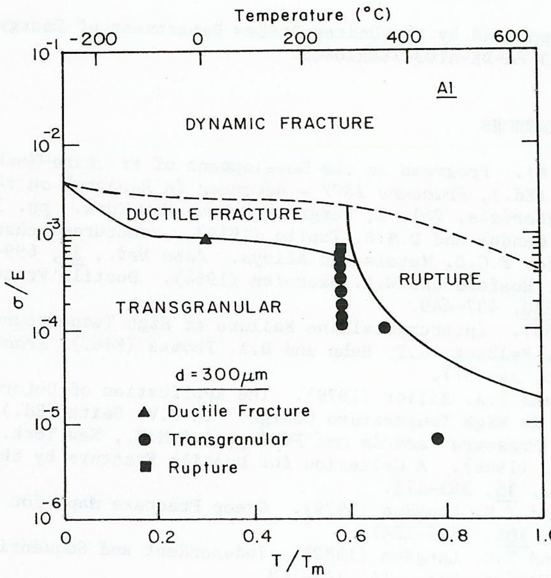

Fig. 5. Experimental fracture map for high purity Al derived by Ashby, Gandhi and Taplin (1979), showing experimental points obtained in this investigation.

several fracture maps derived by Ashby, Gandhi and Taplin (1979) for a number of pure metals and commercial alloys shows no example of a transition from transgranular fracture to intergranular fracture and again to transgranular fracture with increasing σ/E (equivalent to increasing $\dot{\varepsilon}_i$) at constant temperature. Furthermore, this type of behavior would not be predicted by a numerical evaluation of the times to fracture for the different processes (Miller and Langdon, 1979).

SUMMARY AND CONCLUSIONS

1. The fracture strain, ε_f, is almost independent of imposed strain rate in the range $\sim 10^{-6} - 10^{-2}$ s^{-1} for high purity aluminum at 573 K, whereas the Al-5% Mg solid solution alloy shows two maxima in ε_f at $\sim 10^{-5}$ and $\sim 3 \times 10^{-3}$ s^{-1} at the same temperature.

2. High purity aluminum fails primarily in a transgranular manner over a wide range of strain rates at 573 K. In Al-5% Mg, the two maxima in ε_f at 573 K correspond to transgranular fracture, there is a ductility minimum at $\sim 3 \times 10^{-4}$ s^{-1} due to the advent of intergranular fracture, and at very rapid strain rates there is a transition to knife edge rupture. These transitions in Al-5% Mg give rise to marked differences in the appearance of the specimens after fracture.

3. By combining the experimental data on high purity Al with other observations obtained over a range of temperatures, it is shown that the results are in good agreement with an experimental fracture map. No equivalent map is at present available for Al-5% Mg.

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