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TEMPERATURE DEPENDENCE OF THE IMPACT PROPERTIES OF SOME ALUMINUM ALLOYS CONTAINING LOW MELTING POINT INCLUSIONS

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INTRODUCTION

Alloying additions of low melting point metals are often employed to improve the machining properties of a variety of base metals; lead is added to steels and copper alloys, while lead and bismuth are commonly added to both zinc and aluminum alloys. It has been reported that small additions of cadmium to aluminum alloys will also have a beneficial effect on the machining characteristics [1]. In free-machining aluminum alloys, short discontinuous chips appear to form as soon as the heat generated during cutting melts the inclusions [2], although the reasons for the observed marked effect on chip morphology are not known.

The temperature dependence of the mechanical properties of free-machining leaded steels has been studied by Breyer and co-workers [3, 4, 5]. They found that, between 205°C and the melting point of lead, the reduction of area at fracture decreased rapidly to a minimum which occurred at or just slightly above the melting point, 327°C; a similar trend has been observed in impact tests. This degradation of the mechanical properties appears to be associated with liquid metal embrittlement of the steel by lead, with fracture occurring along prior austenite grain boundaries. The present series of tests on binary, high purity Al-Pb, Al-Bi and Al-Cd alloys were planned to establish if a similar correlation exists between mechanical properties and machining characteristics in these alloys and, if possible, to define more clearly the embrittling mechanism stemming from the presence of liquid inclusions.

As machining is a high strain-rate operation, it was decided to investigate the temperature dependence of impact properties. Impact tests were also carried out at fixed temperatures slightly above the melting point of the inclusions, in order to establish the effect of inclusion volume fraction on the fracture characteristics of the alloys.

EXPERIMENTAL TECHNIQUE

The alloys were prepared by induction melting 99.999% pure Al in a graphite crucible under purified argon. An addition of 0.3 wt.% Ti in the form of an Al + 5 wt.% Ti master alloy was made in order to refine the as-cast grain structure. Finally, the desired amount of Bi, Cd or Pb was added to the melt, which was agitated for a few minutes before casting into a graphite mould. Alloys containing up to 1.5 wt.% Cd, 1.5 wt.% Bi or 3 wt.% Pb were studied.

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The microstructures of the Al-Bi, Al-Cd and Al-Pb alloys were similar, showing spherical or slightly faceted inclusions within the grains and lens-shaped particles along the grain boundaries (Figure 1). Inclusion sizes ranged from 0.1 to 3 µm and grain sizes measured by the linear intercept method were in the 0.1 - 0.2 mm range.

The impact tests were carried out with a Hounsfield Impact Tester (impact energy: 65J, impact velocity: 6.9 m/sec.) The specimens were preheated to 500°C in a furnace before being quickly positioned in the machine and allowed to cool to the desired temperature for testing. The specimen temperature was monitored by a thermocouple attached to the notch of the impact specimen, and continuously recorded. The fracture surfaces were observed using scanning electron microscopy.

RESULTS

The energy absorbed during the impact testing of the Al-Bi, Al-Cd and Al-Pb alloys drops suddenly in a narrow temperature range, being fairly constant both above and below the transition temperature (see Figure 2). The curves shown in the figure refer to representative compositions of each of the three alloy systems considered. A classical ductile to brittle transition is indicated by the results; below the transition temperature the specimens were bent without breaking, while above it, in the case of Al-Bi and Al-Cd alloys at least, they fracture with little associated plastic deformation. An Al-Pb alloy containing 3 wt.% Pb never fractured completely when the inclusions were molten, the crack propagating only half way through the specimen. Specimens containing a greater volume fraction of Pb inclusions may be completely brittle, but this is difficult to prove as Al-Pb alloys containing a suitably fine dispersion of inclusions cannot be prepared by normal casting methods in alloys containing more than 1.5 wt.% Pb, the result of a miscibility gap in the liquid state. The transition temperatures in the Al-Cd and Al-Pb alloys were found to occur close to the respective melting points of Cd and Pb. In the case of Al-Bi alloys the transition temperature was found to be about 40°C below the melting point of Bi; this is related to the large supercoolings that are found for Bi inclusions in this alloy [6].

Examination of the fracture surfaces for all alloys showed that the fracture path was intergranular as illustrated in Figure 3 for the case of an Al + 0.1 wt.% Bi + 0.3 wt.% Ti alloy. At this relatively low concentration of Bi, there was still evidence of localized plastic deformation at some locations on the fracture surface.

The energy absorbed during impact was also measured as a function of the concentrations of Bi, Cd and Pb at a constant temperature of 20°C above the respective melting points of the inclusions. Low concentrations of Cd, and particularly of Bi, are sufficient to cause brittle fracture; on the other hand, an Al-Pb alloy containing 1.5 wt.% Pb retains some ductility. It is possible to compare the efficiencies of the three types of inclusions in promoting brittleness in terms of their volume fractions in the microstructure. The results indicated that embrittlement set in at inclusion volume % fractions of 0.007% and 0.06% in Al-Bi and Al-Cd respectively, while Al-Pb is not completely embrittled at a volume % fraction of 0.72%. However, such comparisons can be misleading, particularly if the distributions of the inclusions is non-uniform, as is likely in a cast alloy. The actual coverage of grain boundaries by inclusions was determined by intergranular decohesion by liquid gallium [7];

Figure 4 shows an example of the grain boundary coverage in an Al-0.2 wt.% Cd-0.3 wt.% Ti alloy. Measurements from areas such as are shown in Figure 4 revealed that the local volume fraction of inclusions at the grain boundary was 10 to 20 times greater than what would have been expected from a homogeneous distribution.

DISCUSSION

To gain a better understanding of the origin of the ductile to brittle transition, and of the different effects of Pb on one hand, and of Cd and Bi on the other, the fracture process may be considered in two steps, viz. crack initiation and crack propagation.

With regard to crack initiation, there appears to be two quite distinct ways in which a liquid inclusion could influence the fracture site, depending upon whether the specimen has reached general yielding or not. In the case where general yielding has not been reached, a liquid inclusion can act to concentrate the local *elastic* strain field. This stress concentration will be greatest for the lens-shaped inclusions lying at the grain boundaries. An exact elastic calculation for this inclusion shape is very difficult, but a rough idea of the magnitude of the effect can be made by approximating the lens shape by an oblate spheroid (an ellipsoid having a = b > c, where 2a, 2b and 2c are the lengths of the three principal axes of the ellipsoid). In this case, the stress distribution at the inclusion-matrix interface (see Figure 5) can be found for different values of the applied stress field (p_{ij}^A) following Eshelby's method [8].

The results of this calculation have shown that:
(a) the stress distribution at the surface of a liquid Pb oblate spheroid inclusion varies only slowly with the c/a ratio, e.g., in the case of an applied stress p_{33}^A (parallel to the c-axis of the inclusion - Figure 5), the stress concentration p_{33}^C at the edge of the inclusion parallel to c is 1.94, 2.5 and 3.5 times p_{33}^A when c/a is 1, 1/2 and 0 respectively.
(b) the stress concentration is not very sensitive to the exact form chosen for p_{ij}^A, although the hydrostatic component of p_{ij}^A does have a marked effect on plastic yielding or fracture at the inclusion (see below).

Since the stress distribution does not depend strongly upon the c/a ratio, the results of these calculations should be good approximations to the actual stress concentrations at grain boundary inclusions. As the dihedral angles of liquid Bi, Cd and Pb inclusions in Al are similar (mean values were measured as 105°, 107° and 118° for Al-Bi, Al-Cd and Al-Pb respectively at 350°C), then the c/a ratios vary between .50 to .57, assuming the true grain boundary lens shape consists of double hemispherical caps. Since the compressibilities of the three inclusion materials do not differ greatly (Bi: $4.2 \times 10^{-11}\ N^{-1}\ m^2$, Cd: $3.2 \times 10^{-11}\ N^{-1}\ m^2$, Pb: $3.5 \times 10^{-11}\ N^{-1}\ m^2$), the stress distributions of grain boundary inclusions in all three alloys will be quite similar.

As the tensile load is increased, the stress concentration at an inclusion (c/a = 0.5) lying at the grain boundary can be relieved by plastic flow of the surrounding Al matrix, by decohesion of the matrix - inclusion interface or by grain boundary decohesion. For a simple tensile stress p_{33}^A, it was found that the macroscopic yield stress (as determined from the Von Mises criterion) is reached at the inclusion - matrix interface when the stress concentration is $1.2\sigma_{ys}$, where σ_{ys} is the yield stress of the Al matrix. If instead the inclusion is located close to the root of a blunt notch (e.g., in an impact specimen), where a state of plane strain exists, there is a large hydrostatic component in p_{ij}^{A}. An approximate solution for this case indicates that local yielding takes place when the stress concentration is $2.5\sigma_{ys}$, and the stress component normal to the interface (p_{11}^{C} in Figure 5) is $1.4\sigma_{ys}$. Fracture of either the grain boundary or the inclusion - matrix interface should be more favoured in a Charpy test than in a tensile test, but whether or not fracture does occur prior to plastic flow depends upon the relative strengths of the matrix - inclusion interface and of the grain boundary. One possible explanation for the increased embrittlement observed for Cd and Bi inclusions, but not for Pb, is a marked lowering of the grain boundary fracture strength in the presence of Cd or Bi [9].

A second quite different way in which liquid inclusions could initiate fracture in an impact test is similar to that proposed to explain grain boundary fracture in precipitation hardened alloys [10]. In this model, after plastic flow has occurred in the matrix, dislocation pile-ups form at the grain boundaries leading to fracture of the brittle intermetallic precipitate particles there and localized ductile grain boundary fracture. The stress concentration in this case is caused by the dislocation pile-ups and the local plastic incompatibility of matrix and inclusion. For a liquid inclusion there is no plastic incompatibility, the inclusion adopting (by fluid flow) any shape demanded by plastic flow in the matrix. The grain boundaries, however, act as dislocation barriers so that there is still a higher stress at the boundaries. A propensity to grain boundary fracture would then depend on the inherent strength of the grain boundary in the presence of Cd, Bi or Pb, as well as testing variables such as strain rate and temperature. A high strain-rate test on a notched bar (such as the Hounsfield specimens) would favour brittle fracture.

Grain boundary cracks were initiated in all three alloys studied in this investigation, but only in Al-Bi and Al-Cd was a complete intergranular fracture observed. Similar effects in leaded steels were explained in terms of liquid metal embrittlement [3, 4, 5]. The most generally accepted mechanism of liquid metal embrittlement is the "adsorption-induced, reduction in cohesion" model [11], where an atom of the liquid metal is adsorbed at the crack tip reducing the binding strength of the atoms there and allowing crack propagation at a lower stress level [11]. A similar view of the embrittlement mechanism, and one which is more amenable to numerical calculation, considers the inherent reduction in the grain boundary fracture energy if liquid (or atoms of the liquid inclusion species) are present at the crack tip. In this case, the work of fracture, assuming a completely brittle fracture, becomes $2\gamma_{SL} - \gamma_{gb}$ instead of $2\gamma_{SV} - \gamma_{gb}$ (where the γ's are the appropriate surface energies), representing a reduction in fracture energy of about an order of magnitude. In the present alloys, the propagation of a brittle crack could be explained by liquid metal embrittlement of Al by Cd and Bi, and to a lesser extent by Pb, the grain boundary liquid inclusions supplying the embrittling atoms to the crack tip by vapour transport or surface diffusion. At very low Bi or Cd levels, some measure of ductility is retained, although the fracture path is still intergranular. As the volume fraction of liquid inclusions decreases, the local area fraction of inclusions at the grain boundaries becomes extremely variable, and some grain boundaries may have almost no inclusions. In areas such as these, there appears to be some plastic work associated with fracture as local areas of ductile tearing are visible on the fracture surface. If surface diffusion or vapour transport controls the rate at which the embrittling atoms reach the crack tip, this result could be explained by kinetic factors, for in areas of low inclusion

volume fraction, Cd or Bi have to be transported over increasingly large distances during crack growth.

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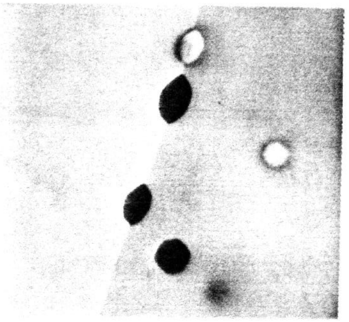


Figure 1 T.E.M. Showing Pb Inclusions at a Grain Boundary in an Al + 1.5 wt% Pb Alloy (X 10700)

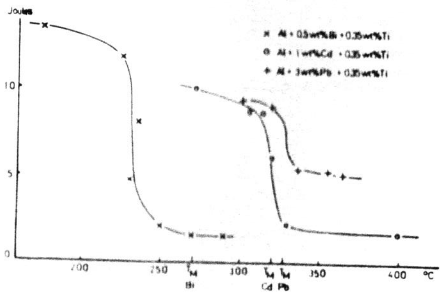


Figure 2 Temperature Dependence of the Energy Absorbed During a Hounsfield Impact Test of Al + 0.5 wt% Bi + 0.35 wt% Ti, Al + 1 Wt% Cd + 0.35 wt% Ti and Al + 3 wt% Pb + 0.35 wt% Ti

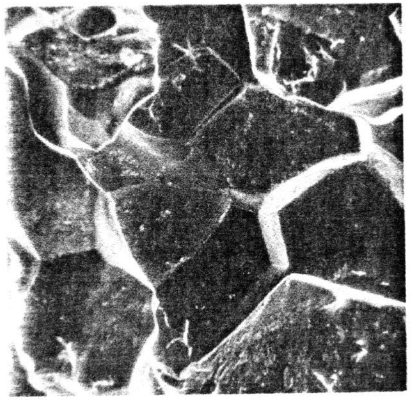
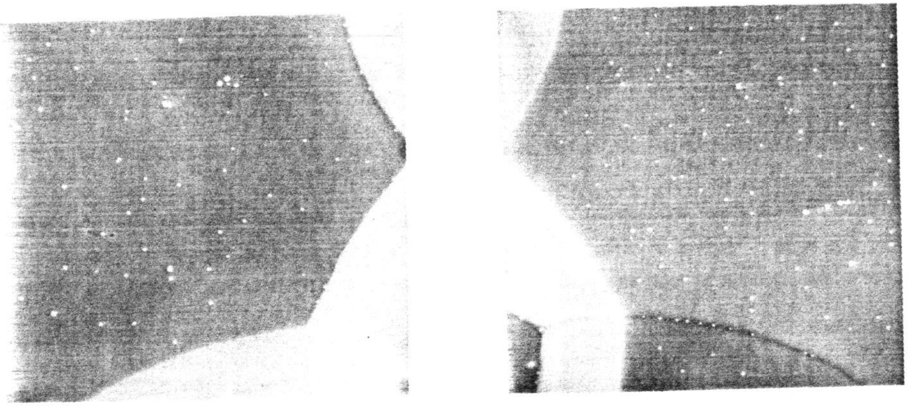


Figure 3 Scanning Electron Micrograph of the Fracture Surface of Al + 0.1 wt% Bi + 0.3 wt% Ti, Impact Tested at 290° C, Showing an Intergranular Fracture (X 250)



(a) (b)

Figure 4 Scanning Electron Micrographs of Matching Grain Boundary Areas of Al + 0.2 wt% Cd + 0.3 wt% Ti

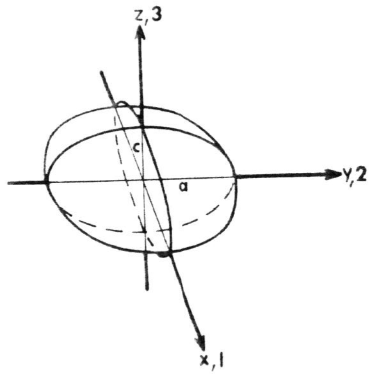


Figure 5 Oblate Spheroid Model of an Inclusion Lying Along a Grain Boundary