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FINAL TECHNICAL REPORT

CORRELATION BETWEEN ACOUSTIC EMISSION AND MICROSTRUCTURE

by

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ABSTRACT

Acoustic Emission from pure metals undergoing deformation is caused by dislocation unpinning and can be described by a theory which was a modified version of Gilman's mobile dislocation density theory. Acoustic Emission from alloys can be dislocation related, but is primarily due to inclusion fracture. Factors affecting this include inclusion size, inclusion density, and the stress state. Inclusions crack more frequently during tensile testing than during fracture toughness testing.

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NOMENCLATURE

E - total energy released
 $\bar{n}$ - number of unpinning events which have occurred
 $\bar{L}$ - average dislocation length
 E_{RMS} - energy to the RMS meter
 RMS - root mean squared voltage
 T - time
 K - constant
 $\dot{\epsilon}_p$ - strain rate
 ϵ_p - true plastic strain
 V - specimen gauge volume
 Φ - total dislocation density
 ρ_m - mobile dislocation density
 ρ_0 - number of initially mobile dislocations
 ϕ - constant
 A - constant
 n - constant
 D - constant
 E - constant
 C - constant
 n' - strain hardening exponent
 σ_{tr} - true stress
 A' - constant
 m - constant
 R - constant

I. Introduction

Acoustic Emission (AE), as defined by the Acoustic Emission Working Group, is "the class of phenomena whereby transient elastic waves are generated by the rapid release of energy from a localized source or sources within a material..." During deformation or fracture of metals and alloys, AE can be attributed to a number of sources, and, in fact, the total observed response from a material is usually a combination of signals from a number of sources.

In pure metals subjected to plastic deformation, a number of authors have suggested possible mechanisms. These include twinning (1), Luder's band formation (2), dislocation multiplication (3,4), pileup and breakaway of dislocations (3,4), dislocation unpinning (5), and slip advances (6). In alloys subjected to deformation, AE may be partially dislocation related, but is, in general, primarily related to cracking or decohesion of inclusions within the matrix. During fracture, AE may be attributed to dislocation activity due to plastic deformation of the matrix around the crack tip, matrix cracking, and/or cracking of inclusions within the matrix.

This investigation was undertaken to identify the source of AE in a number of pure metals and three common aluminum alloys and to correlate these results with the existing literature. In order to obtain our goals, the project was divided into three parts. Initially, the effect of large inclusion cracking on AE during fracture toughness testing of 7075-T651 aluminum was studied. Next, AE data for pure Al, Ni, Cu, and Zr was used to check a theory by Hamstad and Mukherjee (5) relating AE to a dislocation pinning process. Finally, the AE responses of 2124-T851 and 2048-T851 were compared. These responses were also correlated with each material's

fracture behavior and microstructure. For convenience to the reader, each phase of the project will be presented separately.

II. AE From Large Inclusions During Plastic Deformation and Fracture of 7075-T651 Aluminum.

Introduction

Bianchetti et. al. (9) has shown that by monitoring the AE during a tensile test of 7075-T651 aluminum, the presence of large 45 to 120 μm intermetallic inclusions can be detected. That is, if the large inclusions are present, the typical smooth continuous AE curve will have large bursts superimposed upon it. When the 45 to 120 μm inclusions are not present, Graham et. al. (10) and Carpenter et. al. (11) have related the AE response to the fracture of 2-5 μm inclusions within the aluminum matrix.

During fracture toughness testing, burst emissions are observed whether or not the large inclusions are present. Since the burst emission generated by various cracking processes is indistinguishable from that generated by the inclusions cracking, AE reveals nothing about the degree of cracking of inclusions during the fracture toughness test.

The following summarizes the findings of the part of the project which was undertaken to complete the investigation (originally began by Bianchetti) of the properties of the 45 to 120 μm inclusions in 7075-T651 aluminum and to determine the extent of particle fracture during both tensile and fracture toughness testing. Specifically, the size, shape, distribution and extent of cracking of the large inclusions were determined during the investigation.

EXPERIMENTAL PROCEDURE

Tensile samples of dimensions documented in Reference 9 were cut from a 12.7 mm (0.5 inch) plate of 7075-T651 aluminum. These were then strained 10% and sectioned for metallographic observation. Sample

Figure 1: Orientations of tensile and fracture toughness specimens. Metallographic specimens were cut from the gauge length, plastic zone, and the untested regions so that the WT, RT and WR planes could be observed.

Figure 2: Fracture toughness specimen dimensions (mm)

orientations are shown in Figure 1.

Fracture toughness specimens of dimensions shown in Figure 2 were cut from the same plate of aluminum. These specimens were annealed to increase the plastic zone size after testing and were strained until a large dimple indicating the plastically deformed region developed near the crack tip (Figure 3). Metallographic samples of orientations shown in Figure 1 were cut from both the plastically deformed region and the unstrained regions of the sample.

All samples were then mechanically polished using standard techniques, etched with Keller's reagent, and observed with a Bausch and Lomb metallograph. The procedure of polishing, etching, and observing was repeated until the sample had been depleted. Final data consisted of approximately 150 optical micrographs of individual inclusions.

RESULTS

By sectioning many inclusions and observing them in two perpendicular planes it was revealed that the large inclusions are 45-120 μm in all dimensions. Figure 4 shows views of a typical inclusion after sectioning. It was also observed that the large inclusions are concentrated within small areas of the sample and that these inclusions tend to be aligned as shown in Figure 5. Additionally, the major axis of individual inclusions tends to be aligned with the sample rolling direction.

Approximately 40% of the large inclusions within the untested region of the sample contained single fine cracks as shown in Figure 6. These cracks were usually extremely straight and showed no preferred orientation with respect to either the rolling direction or the inclusions' major axis.

Many of the inclusions found within the plastic zone of the fracture toughness samples contained multiple wide cracks ranging in width from

Figure 3: Polished sample showing a dimple due to plastic deformation. The dimple indicates the boundaries of the plastic zone on the sample's surface.

Figure 4: Two perpendicular views of an inclusion

Figure 5: An example of inclusion alignment

Figure 6: An untested inclusion (Rolling direction is horizontal).

0.5 to 3.0 μm . An example of this is shown in Figure 7. Narrow cracks similar to those in the untested regions were also observed. As before, the narrow cracks seemed to show no preferred orientation. However the wider cracks show a definite tendency to align themselves perpendicular to the major axis of the inclusions. Note that this also makes the wide cracks perpendicular to the rolling direction.

All inclusions within the tensile specimens contained multiple large cracks ranging in width from 0.5 to 3 μm . These cracks were aligned perpendicular to the inclusions' major axis, to the sample rolling direction, and to the loading direction. Some fine cracks were also observed. Figure 8 shows an example of an inclusion within the gauge length of a tensile sample.

DISCUSSION

A previous paper (9) has shown that the large inclusions in the 7075-T651 plate were formed during solidification. During the course of this investigation, it was observed that the inclusions are segregated within small areas of the matrix and that on many occasions they are aligned with each other. It is these two observations that led to the following possible explanation of why the large inclusions are present in the aluminum plate.

Prior to casting the molten aluminum must be transferred from a holding furnace to the casting head. This transfer must take place very smoothly and is done by means of a Launder system (12). It is usual practice to include a filtration system consisting of either one or two woven glasscloth filter bags or a sintered alumina filter bed in the Launder system. In the case of the 7075-T651 plate in question, the filtering system could have been defective since this would not only allow

Figure 7: An inclusion within the fracture toughness plastic zone (Rolling direction is horizontal).

Figure 8: An inclusion from the tensile gauge length
(Rolling direction in horizontal).

the solid chromium-rich inclusions to pass through the filter but would also cause the particles to become aligned as they pass through. Since flow between the filter and the casting head is smooth, we would expect the inclusions to be segregated or aligned within the matrix. Experimental observations support this.

If the inclusions from the three regions are compared, it is found that the inclusions from the untested samples contain both the fewest and the least severe cracks, the inclusions from the plastic zone contain more cracks, and the inclusions from the tensile sample contain the most cracks. Since this is the case and since the larger cracks in the strained samples show a preferential orientation, it can definitely be stated that new cracks nucleated during the testing process. The cause of this new crack nucleation cannot be determined from experimental evidence since both the major axis of these inclusions and the sample rolling direction coincide in the fracture toughness sample, and the major axis, rolling, and loading direction coincide in the tensile sample. That is, since the minimum area of the inclusions lies perpendicular to the rolling direction, it is not obvious whether the preferential crack orientation is due to stress variation in the matrix or to geometrical considerations.

As was discussed in Reference 9, the burst type emission observed during tensile tests is due to the fracture of large inclusions. Since inclusions cracking also occurs during fracture toughness testing some of the burst emission which is observed probably comes from particle cracking. Since the percentage of cracked inclusions within the plastic zone is variable due to segregation the amount of contribution of cracking is difficult to predict. However, since many bursts occur during the test due to other mechanisms and few inclusions are seen within the

plastic zone, the AE response from aluminum with the large inclusions is not markedly different from the response of aluminum without inclusions.

III. Acoustic Emission from Pure Metals

Introduction

A "pure" metal, for the purposes of this discussion, is a metal or alloy which would contain no inclusions which could crack or decohere to cause AE. That is a pure alpha brass would be considered in much the same way as would pure Cu or Zn.

A number of authors have studied the AE responses from pure metals and have concluded that the mechanism involved is related to some type of dislocation activity. Many reasons have led to these conclusions. One of the most convincing of these can be reviewed by comparing the AE response (as measured by a true RMS voltmeter) with a materials stress strain curve. Generally, materials such as Cu, Ni, or Ag with smooth stress strain curves exhibit smooth continuous AE curves. Usually, the RMS level from the material remains low until just before yield (when the longer dislocation segments begin to move) reaches a maximum at approximately yield (probably due to an increase in dislocation activity) and decays slowly with increasing strain. These correlations and the fact that the RMS curve resembles Gilman's curve for mobile dislocation density was one of the first indications that AE was dislocation related.

In iron alloys and alpha brasses which exhibit discontinuous deformation characteristics, AE shows large increases when dislocation motion increases thus again implying that AE is dislocation related. The observations that AE decreases with increasing cold work (14), decreasing grain size (15), and neutron irradiation (16) also support the theories of AE being dislocation related.

Even though most authors agree that AE from pure metals is dislocation related, few agree on the specific process which is causing the

emission. Some of the suggestions of possible mechanisms which have been made include dislocation multiplication, pile-up and breakaway of dislocation tangles, dislocation unpinning, initial dislocation motion, and slip advances. Of these, the experimental data for pure metals which deform continuously (that is, those which have a smooth stress strain curve) support a dislocation unpinning mechanism.

Some examples of results which support this are given by Schofield (17) working with gold in 1964, James and Carpenter (18) working with LiF, Carpenter, et. al. (11) working with the first dislocation peak in 7075-T6 aluminum and Olson, et. al. (13) working with Al-Cu alloys. Carpenter, as well as Olson, using internal friction technique, has shown that the AE peak corresponds to an internal friction peak caused by dislocation interaction with pinning points whether these are solute atoms or precipitates. Frederick (14) has observed that the AE from a metal is decreased by a decrease in initial dislocation segment length due to cold work or precipitation hardening. Nakasa has observed this same effect by studying the effects of neutron irradiation on AE in pure copper. Since the energy released during unpinning is related to the segment length involved, this supports an unpinning theory. Final support for the dislocation pinning theories comes from Eisenblatter (19), who has successfully correlated the AE response in the form of an RMS voltage to the mean free path of a dislocation within the material. This, in fact, would mean that pinning must be an important factor effecting the AE response.

The following is a summary of the phase of this project which was undertaken to check a dislocation unpinning theory by Hamstad and Mukherjee (5) using pure Al, Ni, Zr, and Cu as typical pure metals.

Data for the copper in both the cold worked and annealed states was generated in our laboratory. Data for the Al, Ni, and Zr was obtained from Imada, et. al. (20).

Theoretical Considerations

In a recent paper by Hamstad and Mukherjee (5), a theory relating AE (in the form of RMS from a true RMS voltmeter) to the process of dislocation unpinning was developed. The following is a modified version of their theoretical analysis which will take into account both the initial dislocation density as well as the dislocation density due to plastic deformation.

If we assume that the source of AE in pure materials is due to the energy released when a dislocation unpins, we would expect:

$$E \propto \bar{n} \bar{L}$$

where E is the total energy released, $\bar{n}$ is the number of unpinning events which have occurred, and $\bar{L}$ is an average dislocation length (since all lengths are approximately the same).

By the definition of RMS, we know that the energy to the meter from the RMS signal is:

$$E_{\text{RMS}} \propto \text{RMS}^2 \Delta T$$

RMS is the root mean square voltage, ΔT is time and E_{RMS} is the energy from the signal.

If we assume that all system losses are linear transfer functions, then the total energy from the events is represented by

$$E \propto E_{\text{rms}}$$

After absolute noise subtraction (5), the RMS voltage would be given by the equation

$$\text{rms}^2 \Delta T = K^2 \bar{n} \bar{L} \quad (1)$$

where T is time and K^2 is a constant.

Since we know $\dot{\epsilon}_p = d\epsilon_p / dt$

where ϵ_p is the plastic strain and $\dot{\epsilon}_p$ is the strain rate, Equation 1 can be rewritten as:

$$rms^2 = \frac{K^2 \bar{n} \bar{L} \dot{\epsilon}_p}{\Delta \epsilon_p} \quad (2)$$

If ϕ is the total density of dislocations that have come unpinned up to any plastic strain ϵ_p , then for the increment $\Delta \epsilon_p$

$$\frac{\bar{n} \bar{L}}{\Delta \epsilon_p} = V \frac{\Delta \phi}{\Delta \epsilon_p}$$

where V is the specimen volume. If we now substitute into Equation 2 and take the limit as $\Delta \epsilon_p \rightarrow 0$ we find

$$rms = K V^{1/2} \epsilon_p^{1/2} \left(\frac{d\phi}{d\epsilon_p} \right)^{1/2} \quad (3)$$

Note that the assumption that ϕ is independent of strain rate has been made.

From Gilman's model (21), the mobile dislocation density is represented by the equation:

$$\rho_m = \left(\rho_0 + A \epsilon_p \right)^n e^{-\phi \epsilon_p}$$

where ρ_m is the mobile dislocation density, ϕ , A , n are constants, ϵ_p is plastic strain and ρ_0 is the number of initially mobile dislocations. (Generally, ρ_0 is considered negligible in high strength alloys or after a few percent strain).

If we now assume that $d\phi/d\epsilon_p$ has the same dependence on plastic strain as the mobile dislocation density, then

$$\frac{d\phi}{d\epsilon_p} = (D + E \epsilon_p^n) \exp(-\phi \epsilon_p)$$

where D and E are constants.

Substituting this into Equation 3, we find:

$$RMS^2 = C V \epsilon_p (D + E \epsilon_p^n) \exp(-\phi \epsilon_p) \quad (4)$$

where C is a constant.

This is the theory which will be used to predict theoretical RMS values for the comparisons made in the remainder of the report. Note that the RMS value in Equation 4 is the actual RMS after absolute noise subtraction is used and that the assumptions that the flow stress was not a function of the strain rate was made. Also note that the theory is limited to materials which exhibit a smooth stress-strain curve.

In order to determine the constant n in Equation 4, examine the equations:

$$\sigma_{tr} = A' \epsilon_p^{n'} \quad (5)$$

$$\rho_T = A'' \epsilon_p^n \quad (6)$$

and
$$\sigma_{tr} = D \rho_T^{1/2} \quad (7)$$

where σ_{tr} is the true stress, ρ_T is the total dislocation density, and A' , n' , A , n and D are constants.

Substitute Equation 6 into Equation 7 and square the result to find:

$$\sigma_{tr}^2 = D^2 A''^2 \epsilon_p^{2n} \quad (8)$$

Equation 5 can be squared to show

$$\sigma_{tr}^2 = A'^2 \epsilon_p^{2n'} \quad (9)$$

If Equation 8 is then substituted for σ_{tr} in Equation 9, we find:

$$A'^2 \epsilon_p^{2n'} = D^2 A''^2 \epsilon_p^{2n}$$

Since this is the case: $2n = n'$

n' is the strain hardening exponent which can be determined from the materials stress strain curve.

No other constants in Equation 4 could be easily derived from first principles.

Experimental Procedure

Tensile samples of dimensions shown in Figure 9 were machined from a 12.5 mm (0.5 inch) copper plate of 99.9% minimum purity. Part of

Figure 9: Copper tensile sample dimensions (mm)

these were annealed enough to cause full recrystallization. All were then stored in the atmosphere at ambient temperature until testing began.

The AE response of each sample during a tensile test was measured using a model TTD Instron testing machine and an AET Model 201 Acoustic Emission Laboratory Signal Processor. The Signal Processor consists of a model AC175L resonant type PZT transducer with resonance at 175Khz, a band pass filter set at 125-250Khz, a preamplifier (69db gain), an amplifier (32db gain) and a true RMS volt meter. A block diagram of the equipment is shown in Figure 10. Note that the transducer is coupled to the sample with AET SC3 couplant and is held in place with a rubber band.

The procedure followed during testing included a number of steps to eliminate the effects of extraneous noise and to remove anything which could attenuate the true AE signal. Initially, the sample pin areas were preloaded to 1.3 times the ultimate strength of the gage length so as to avoid AE being generated from the pin areas. Specimens were then etched with a dilute nitric acid solution prior to testing to remove surface oxides. Teflon plates were placed between the fixtures of the Instron and between the loading pins and gripping fixtures to reduce extraneous noise. Finally, the 25.4 mm (1 inch) gage length extensometer for strain measurement, and the transducer were attached to the sample, and a light coating of grease was applied just outside the gage length to decrease noise effects.

All tests were run at a strain rate of $1.67 \times 10^{-3} \text{ sec}^{-1}$ with noise levels being measured before and after each test. Fracture surfaces and surface textures were visually examined after each test. The final data consisted of Force - Elongation and RMS - Elongation curves for copper in both the annealed and cold worked state. These were then converted to True Stress - Strain and RMS - True Plastic Strain curves.

Figure 10: AE apparatus block diagram

Stress-Strain and RMS-Strain curves for Ni, Al, and Zr were obtained from the literature (20). Table 1 summarizes the important parameters used during the tests. Complete details of the experimental apparatus and procedure used to generate this data can be found in Imaeda (20).

It should be noted that all RMS values tabulated hereafter have had noise levels removed using an absolute subtraction technique (5).

Results

Experimental results for Cu, Al, Ni and Zr were compared with theoretical predictions from the equation:

$$\text{RMS}^2 = C V \epsilon_p (D + E \epsilon_p^n) \exp(-\phi \epsilon_p)$$

with n being determined from the strain hardening exponent of each material. Since no reasonable correlations could be made, the theoretical equation was modified so that:

$$\text{RMS}^2 = C V \epsilon_p [D \epsilon_p^m \exp(-R \epsilon_p) + E \epsilon_p^n \exp(-\phi \epsilon_p)] \quad (10)$$

m and R are constants.

This equation uses the part of Gilman's theory (21) which is used in most experimental applications. However, the modifications allow the initial dislocations to build up and decay at a rate different from those generated during deformation. Figures 11 through 14 show a comparison of experimental results and theoretical predictions using Equation 10.

Figures 11 through 13 for the fcc metals Cu, Ni, and Al respectively show that theoretical predictions are in excellent agreement with the average AE behavior throughout the regions of strain examined. Table 2 gives the equations used to predict the theoretical RMS values. As was expected, the n values for the materials decreases in the order of decreasing stacking fault energy, that is in the order Ni to Al to Cu.

TABLE 1: Experimental Parameters

Material	Purity	Structure	Heat Treatment
Aluminum	99.89%	FCC	400°C x 1 hr FC
Nickel	99.6%	FCC	800°C x 1 hr FC
Zirconium	99.6%	HCP	1000°C x 1 hr FC

FC = furnace cool

GAIN: 80dB for FCC samples
60 dB for HCP samples

STRAIN RATE: $2 \times 10^{-4} \text{ sec}^{-1}$

FILTER: 100 to 600Khz

Figure 11: Comparison of Theoretical predictions and Experimental results for copper.

Figure 12: Comparison of Theoretical predictions and Experimental results for nickel

Figure 13: Comparison of Theoretical predictions and Experimental results for Aluminum

Figure 14: Comparison of Theoretical predictions and Experimental results for Zirconium

TABLE 2: Theoretical Equations Used for Predictions

Material	Theoretical Equation
Copper	$\text{RMS}^2 = 325\epsilon_p \exp(-56.6\epsilon_p) + 82\epsilon_p^{0.5} \exp(-182\epsilon_p)$
Aluminum	$\text{RMS}^2 = 8.76 \times 10^{10} \epsilon_p^3 \exp(-1097\epsilon_p) + 89.8\epsilon_p^{0.78} \exp(-117\epsilon_p)$
Nickel	$\text{RMS}^2 = 7.6 \times 10^5 \epsilon_p^{1.1} \exp(-800\epsilon_p) + 1570\epsilon_p^{1.02} \exp(-163\epsilon_p)$
Zirconium	$\text{RMS}^2 = 9 \times 10^{13} \epsilon_p^{3.6} \exp(-250\epsilon_p) + 1.1 \times 10^6 \epsilon_p^{0.25} \exp(-137\epsilon_p)$

N values given are 1.02, 0.78 and 0.50, respectively. Values for the other constants show no relationship to the stacking fault energy due to the error in strain measurement in the Ni and Al data. (An extensometer was not used for strain measurements of Ni and Al).

Figure 14 compares theoretical and experimental values for zirconium. In this case, although theoretical and average experimental values correlate well, it must be noticed that unlike the results for the fcc metals, there is considerable experimental spread in the data. This allows a wide variation in theoretical constants.

Figure 15 compares AE responses in the form of RMS for annealed and cold worked copper. Two features of this figure need be noticed. First, the peak magnitude for the annealed copper is roughly five times greater than that for the cold worked copper. Secondly, the elastic limit as measured from a stress strain curve occurs at the RMS peak in the cold worked copper, but occurs much before the peak of the annealed copper.

Discussion

As was mentioned previously, a number of authors have attributed AE in pure materials to the unpinning of dislocation during plastic deformation. Also a number of authors (see James (22) for references) have found that the mobile dislocation density in a material is difficult if not impossible to measure. If the Gilman theory:

$$\rho_m = (\rho_o + A \epsilon_p^n) e^{-\phi \epsilon_p} \quad (11)$$

is examined, one finds that the usual assumption that is made is that after a few percent strain $\rho_o e^{-\phi \epsilon_p}$ is essentially zero and can be ignored. In the case of AE, it is the $\rho_o e^{-\phi \epsilon_p}$ which is most important since the RMS peak occurs near yield.

Figure 15: Comparison of the AE responses from annealed and cold worked copper.

Since this part of Gilman's theory is difficult to check and since the AE data warrants such a change, we therefore feel that a slight modification of Equation 11 to:

$$\rho_m = (\rho_o \epsilon_p^m e^{-\phi_1 \epsilon_p}) + A \epsilon_p^m e^{-\phi \epsilon} \quad \phi_1 \text{ and } m \text{ are constants}$$

would be reasonable. In this case the second term on the right hand side of the equation refers to the Gilman theory usually checked by those measuring density by other means and the first term on the right refers to the buildup and decay of the dislocations initially present in the material. Notice that the modification is based on the assumption that the initial dislocations would follow a buildup and decay equation of the same form as the generated dislocations. Only the rate of buildup and decay would be different. Qualitative support for Equation 12 can be found in the literature. Olson, et. al. (13) shows that immediately prior to yield in an AlCu alloy, internal friction increases drastically then rapidly decreases. This friction peak has been related to the interaction of dislocations with solute atoms. Shimizu, et. al. (23) using ultrasonic techniques to study Al and Cu has also shown that the AE peak occurs as the mobile dislocation density increases with segment length held constant. Since the AE peak rises rapidly at approximately yield, this would suggest that Gilman's theory should be modified.

Once the Gilman theory is modified, the AE dependence on each variable in the equation needs to be checked. A number of authors including Shimizu, et. al. (23), Eisenblatter, et. al. (19), and Imaeda, et. al. (20) have confirmed that Equation 10 correctly predicts the dependence of RMS on strain rate. Each has independently conducted experiments and has found that RMS is indeed dependent on strain rate to the one half power for a number of materials including brass, Cu, Zn, Al, Fe, Sn, Ni, and Zr.

Data from the results section of this paper has shown that the strain dependence of Equation 10 fits the data reasonably well. However, in comparing actual values given in Table 2 for the constants n and m,

a number of things must be kept in mind. n has been determined from the stress strain curve and thus is somewhat independent of small errors in strain measurement. However, m has been empirically determined from the RMS curve and has been found to be highly dependent on small errors made during strain measurement. In the case of the experimental data, the curves for Ni and Al were generated without measuring strain with an extensometer, thus m values should not be considered exact. In fact, these m values are so variable that it would seem possible, if strain measurements were entirely accurate, to eliminate the strain dependence of the part of Equation 10 referring to the dislocations initially present within the material. If this could be done, the only modification of Gilman's theory which needs be made deals with the different decay rates of the initial and generated dislocations.

By observing Figures 11-14, a drastic difference in the data spread of the fcc and hcp metals can be seen. In the case of the fcc metals, the AE curve is well defined at all values of strain. On the other hand, the Zr curve shows a large spread in the data. An explanation of this lies in the fact that during deformation, two processes which generate AE occur within Zr. First, dislocation unpinning occurs much the same as it did in fcc metals. Second twinning occurs (this does not occur in the fcc metals examined). Since this is the case, the AE curve observed in the case of Zr is a composite curve consisting of a smooth AE curve due to dislocation motion with burst emissions due to twinning superimposed on it. These bursts cause the data spread.

The effects of cold work on a materials' AE response is illustrated in Figure 15. The observation that the cold worked sample shows a decrease in peak amplitude by a factor of 5 can be explained by considering

the effects of cold work on the mobile dislocation density. Since cold work increases the total density of dislocations, the mobile dislocation density should decrease due to an increase in the amount of dislocation interactions. Since the mobile density is decreased, fewer dislocations can participate in the pinning process. This process is the source of AE so the AE peak is decreased. Note that Equation 10 would predict this decrease.

IV. Relationships Between Fracture Toughness and AE in 2124-T851 and 2048-T851 Aluminum

Introduction

High strength aluminum alloys fracture by a void coalescence mechanism. Initially, the process begins when large second phase particles begin to fracture or decohere from the matrix causing large voids to develop. Since these voids are separated by large distances, void coalescence can not occur. Instead, as the process continues the smaller inclusions begin participating in the void formation process. Eventually the spacing between voids reaches a critical value and coalescence occurs. This completes the fracture process.

Hahn, et. al. (24), Low, et. al. (25) and Ryder (26) have independently summarized some of the variables which affect fracture by void formation and coalescence. Their conclusions are as follows. Increased size and volume fraction of inclusions will generally decrease the fracture strength since larger inclusions usually fracture at lower stresses and increased volume fraction affects void coalescence. Particle properties such as strength and elastic modulus influence fracture in that they control the stress at which the inclusions will break and the stress felt by the inclusions. Interface properties are important since they control decohesion. Local stress concentrations within the matrix and grain size are important. Also, previous processing may have an effect since this may precrack the inclusions. In the case of high strength aluminum, the factors dealing with inclusions are the usual means of raising a material's fracture toughness.

Generally, when considering the effects of inclusions on the fracture behavior of a high strength aluminum, three types of inclusions must be considered (24). The largest inclusions found within the large dimples of the aluminum's fracture surface is approximately 0.1 to 10 μm in diameter and contains iron, copper, and silicon. The iron and silicon which causes these inclusions to form is present as impurities within the matrix, however the copper is an alloying element added to improve aging. The intermediate size inclusions which are generally found within the smaller dimples of the fracture surface generally contain chromium, magnesium, or zirconium. These elements are alloying elements that were added to control recrystallization. The final inclusion within the matrix is a hardening precipitate which is primarily added to strengthen the matrix and plays little role in the fracture process.

Of the high strength aluminums, the 2000 series has been used for a number of aircraft and other engineering construction applications where weight is a major concern. The first Al-Cu-Mg (2000 series) alloy was discovered by Wilm approximately 60 years ago and is now designated 2017 (29). Some years later, metallurgists began to add silicon to the alloy to enhance the artificial aging process and alloy 2014 was born. In the 1930's, 2024 was introduced as a higher strength alloy to replace 2017. Later an alloy, 2124, was developed which had even better mechanical properties than 2024 and recently an alloy, 2048, has been introduced with even better fracture properties. (Tables 3a and b list mechanical properties and nominal compositions of some of these alloys.)

With the improvements in the strength properties of aluminum alloys, problems with fracture have arisen to such an extent that now alloys must meet certain fracture toughness requirements before they can be used for

TABLE 3a: Approximate Chemical Compositions of the 2000 Series
Aluminum Alloys

Alloy	<u>Wt.-pct Elements</u>							
	<u>Si</u>	<u>Cu</u>	<u>Mn</u>	<u>Mg</u>	<u>Cr</u>	<u>Zn</u>	<u>Fe</u>	<u>Other</u>
2014	0.8	4.4	0.8	0.5	0.1*	--	0.7*	----
2024	0.5*	4.4	0.6	1.5	0.1*	--	0.5*	----
2124	0.2*	4.4	0.6	1.5	0.1*	--	0.3*	0.15 total
2048	0.15*	3.3	0.4	1.5	--	0.25*	0.2*	0.15 total

* indicates maximum values

TABLE 3b: Mechanical Properties of the 2000 Series Alloys

<u>Alloy and Temper</u>	<u>Strength</u>		<u>Fracture Toughness</u>	
	<u>Tensile</u>	<u>Yield</u>	<u>Longitudinal</u>	<u>Short Transverse</u>
2014-T6	480	410	23-26	18-21
2024-T8	480	450	22-27	18-22
2124-T8	490	440	31	25
2048-T8	460	240	37	28

All Strengths are given in MPa.

All Fracture Toughness values have units MPa $\cdot m^{1/2}$

Data was taken from reference 32

certain applications. Since fracture toughness testing is difficult and expensive, new testing methods are being explored. Gerberich, working with iron alloys (27), and Graham, et. al. (10) and Bianchetti, et. al. (9) working with aluminum alloys have shown that the primary source of AE from alloys with second phase particles is due to inclusion cracking or decohesion. This, as Mirabile (28) has stated, is because the AE due to fracture or decohesion of particles is much greater than that due to dislocation motion. Since, then, both fracture toughness and AE are related to the cracking or decohesion of inclusions, AE could possibly be used to predict the fracture behavior of a certain plate of material.

Hamstad, et. al. (29) has shown some correlation between AE and fracture toughness in three plates of 2124 aluminum and has shown how AE varies with orientation. However, in the preliminary investigation no attempt was made to related the AE differences to the plate microstructure. During the course of this investigation, the AE from the same three 2124 plates will be related to their microstructural differences. Also AE behavior, fracture toughness values, and microstructures of 2124 and 2048 aluminum will be compared. Correlations among these variables will be made when possible.

Experimental Procedure

Tensile and compression samples of dimensions shown in Figure 16 were cut from three half inch plates of 2124-T851 aluminum. In this case, the loading axes were always parallel to the plate rolling direction. Chemical compositions and fracture toughness values for each are given in Table 4.

Figure 16: Sample dimensions and orientations used during the Aluminum alloy study. a) Tensile sample b) Compression samples c) Reference directions

TABLE 4: Chemical Compositions and Mechanical Properties of the 2124 and 2048 Plates

<u>Alloy</u>	<u>Cu</u>	<u>Mg</u>	<u>Mn</u>	<u>Cr</u>	<u>Zn</u>	<u>Fe</u>	<u>Si</u>	<u>Long. Tensile Strength</u>	<u>Long. K_{IC}</u>
2124	4.21	1.25	0.30	0.01	0.15	0.23	0.16	487	25.1
2124	3.93	1.48	0.36	.002	0.02	0.15	0.11	484	28.7
2124	3.93	1.24	0.60	.001	0.03	0.08	0.11	473	34.6
2048	2.97	1.29	0.24	----	0.25	0.13	0.07	460	37.0

The Zn composition of the 2048 is the nominal composition. Actual composition was unknown.

Strength values are in MPa.

K_{IC} values are in MPa · m^{1/2}

Tensile and compression samples of the same dimensions were also cut from a half inch plate of 2048-T851 aluminum. In this case, loading axes were oriented in either the longitudinal or the transverse direction of the original plate. Compositions and approximate fracture toughness values are given in Table 4.

Prior to testing all pin areas were preloaded to eliminate possible emission from these areas. Teflon was placed between all fixtures of the testing machine and samples were strained 10%. While deformation was occurring AE was monitored using a system similar to that shown in Figure 10. In this case, however, an AET model C175B transducer was used in conjunction with a band pass filter set at either 100-300Khz or 15-300Khz. Overall system gain was 100db.

After testing, the tensile samples were sectioned and mounted in lucite so that planes perpendicular to the longitudinal, long transverse, and short transverse directions could be examined. Specimens were then mechanically polished using standard techniques and were etched with either Kellers reagent or a dilute nitric acid solution. All were then examined with the scanning electron microscope, the optical microscope, and the electron microprobe.

Raw data consisted of approximately 250 micrographs.

Results

Figure 17 shows stress-strain and RMS-strain curves for the 2048 aluminum in compression for samples of longitudinal and long transverse orientation. (Longitudinal orientation refers to the sample loading axis being parallel to the longitudinal direction of the plate). Notice that although the stress-strain curves are approximately the same, the peak RMS after absolute noise subtraction for the longitudinal orientation is

Figure 17: Orientation effects shown by 2048 in compression. Strain rate = 0.0024/sec. Filter = 100-300 kHz

approximately 2.72 times less than that for the long transverse specimen. The longitudinal peak also occurs at a lower strain than the long transverse peak.

Figure 18 compares longitudinal and long transverse responses in the frequency range 100 to 300Khz for the tensile samples of 2048. Two things need to be noted in this case: 1) Two peaks are observed in both cases, and 2) little or no difference in the first peak is observed. By changing the filter used in Figure 18 to 15-300Khz, filtering effects can be examined (Figure 19). In this case, the 15-100Khz component of frequency causes the long transverse peak to exceed the longitudinal peak by a factor of approximately 1.86 after absolute noise subtraction. Recall that this was also the case in compression when the 100 to 300Khz filter was used.

In order to identify the cause of the observed frequency dependence of AE, the effect of filtering on both the longitudinal and transverse planes were investigated. Results are shown in Figures 20 and 21. In the case of the longitudinally oriented sample little or no effect of filtering is observed. In the case of the long transverse sample (Figure 21) the first peak is decreased by a factor of 1.6 (after absolute noise subtraction) when the 15 to 100Khz frequency band is not considered. This difference probably would not be due to the flow stress differences observed.

Figure 22 shows a comparison of the RMS responses of 2124 and 2048 in tension for longitudinal samples. Although the 2048 shows twin AE peaks, one at approximately 0.8% strain and the other at about 3.5% strain, the 2124 sample curves show only one peak. This peak occurs at approximately 1.3% strain for the medium fracture toughness sample and

Figure 18: Orientation effects shown by 2048 in tension.
 Strain rate = 0.00099/sec.
 Filter = 100-300kHz

Figure 19: Filtering effects shown by 2048 in tension.
Strain rate = 0.00061/sec.
Filter = 15-300kHz

Figure 20: AE response shown by 2048 in longitudinal tension. Strain rate = 0.00061/sec

Figure 21: AE response of 2048 in long transverse tension. Strain rate = 0.00061/sec

Figure 22: Comparison of AE responses of 2124 and 2048
in longitudinal tension. Strain rate = 0.00099
Filter = 100-300kHz

at about 1.8% strain for the low fracture toughness sample. Peak heights also vary among the samples with the second peak of 2048 being about 5.3 times less than that of the medium fracture toughness 2124 and about 4 times less than the low fracture toughness peak. (All values were again determined after absolute noise subtraction was used). The ordering of the RMS peaks with medium fracture toughness being highest, low fracture toughness next, and 2048 lowest is different from the stress-strain curve ordering of decreasing strength being from low to medium fracture toughness to 2048.

Figure 23 shows responses of 2124 aluminum of varying fracture toughness during compression testing. Unlike the tensile responses of the material, two AE peaks, one at .7% strain and the other at 3.5% strain, occur during compression testing. Although flow stress decreases from low to medium to high fracture toughness, the first AE peak decreases in magnitude from medium to high to low fracture toughness. Notice, however, that the second AE peaks decrease in magnitude in the order from low to medium to high fracture toughness.

Figure 24 shows a comparison of 2048 and the 2124 high fracture toughness samples in longitudinal compression. The first peak of 2048 is greater than the 2124 peak by a factor of 2. It also occurs at a slightly lower strain than does the 2124 first peak. Even so, the second peak of 2048 is almost negligible when compared to the 2124 second peak. Recall that the high fracture toughness second peak was the lowest second peak observed in Figure 23.

Microscopic examination has yielded a number of facts about the samples studied. When comparing the 2048 microstructure with the 2124 microstructure, the most obvious difference is that the 2048 inclusions are generally smaller than the 2124 inclusions (Figure 25). Figure 25

Figure 23: AE-response of 2124 in longitudinal compression.
 Strain rate = 0.0024
 Filter = 100-300kHz

Figure 24: Comparison of the AE responses of 2048 and 2124 in longitudinal compression. Strain rate = 0.0024/sec
Filter = 100-300kHz

Figure 25: Typical view of inclusions in the 2048 (top) and the 2124 (bottom) low fracture toughness samples. (Both are WT planes)

like many of the other micrographs also seems to show that the density of inclusions in the 2048 sample is greater than that of the 2124 inclusions. This observation may be real or may be related to the fact that the etching rates vary among the samples. If the observation is real, the density difference is observed only in the LW and WT planes.

When comparing the types of inclusions in 2124 and 2048, another basic difference between the alloys is observed. Although both contain a large dark inclusion (Figure 26), only 2124 samples contain large light colored inclusions. Differences in the matrix copper content may have caused this.

The dark inclusions in 2048 were found to be rich in Cu, Si, and Fe (Figure 27). They have the same amount of Mn, Cr, Zn, and Ti as the matrix and have less Mg than the matrix. Although some of the inclusions are regularly shaped in all planes, most are elongated in the rolling direction in the LW and WT planes and are regularly shaped in the LT plane (See Figure 28). This suggests that most of the inclusions are plate-like in shape. Some stringers also occur in all planes.

Some of the inclusions in 2048 are cracked after testing, although the frequency of cracking seems to be less than in any of the 2124 samples. Data, also, suggests that the regularly shaped inclusions in 2048 crack more frequently than do the elongated ones (See Figures 29 and 30).

In 2124, the dark inclusions are generally either regularly shaped or are stringers and are not elongated like the 2048 inclusions (Figure 31). The inclusions contain a concentration of Cu, Mn, Fe, and Si and have the same amount of Zn, Cr, and Ti as the matrix (Figure 32). These, too, are depleted of Mg. Cracks which are observed within these specimens

Figure 26: Types of inclusions in 2048 (top) and 2124 (bottom). (Rolling direction is vertical in the top micrograph and is perpendicular to the plane of the bottom micrograph.)

Cu

Si

Fe

Mg

Mn

Al

Figure 27: Microprobe analysis of an inclusion in 2048.
(1cm= 5microns)

Figure 28: Inclusion shape on each plane of 2048

Figure 29: Typical inclusions and degrees of cracking in 2048-T851. (SEM X2400) (Rolling direction is vertical in the top micrograph and is perpendicular to the plane in the bottom micrograph.)

Figure 30: Typical inclusions and degrees of cracking in 2124-T851. (SEM X1300) (Rolling direction is vertical) Arrows delineate cracks.

Figure 31: Typical inclusions in the 2124 medium
fracture toughness sample.(WT planes are shown)

Cu

Si

Fe

Mg

Mn

Al

Figure 32: Microprobe analysis of a dark inclusion in 2124. (1cm= 5 microns)

are generally perpendicular to the rolling direction and are independent of the inclusion major axis in many cases. Usually the inclusions contain at most a few cracks.

The light inclusion in 2124 is generally more ductile than the dark inclusion in that the frequency of cracking of the light inclusions is much less than that of the dark ones. Light inclusions tend to overlap the dark ones and many times will appear to be connected to them. Even when this occurs, the frequency of cracking of the dark areas exceeds that of the white areas (Figure 33).

The light inclusion is rich only in Cu and contains the same composition as the matrix in Mn, Cr, Zn, Fe, Si, and Ti. It, too, is depleted of Mg (Figure 34).

Variations in microstructure among the 2124 samples is, as expected, much less than that between the 2124 and the 2048 samples. In fact, the differences are so subtle that only two were observable. These are: 1) both the light and dark inclusions in the medium fracture toughness sample appear to be slightly smaller than in either the low or high fracture toughness sample (Figures 35 and 36) and 2) the volume fraction of inclusions in the high fracture toughness sample is less than in the medium or low fracture toughness samples.

Figure 37 showed the grain structure seen in both 2124 and 2048. Both materials seem to have the same appearance with grains being narrow and elongated in the LW and WT planes and regularly shaped in the LT or rolling plane. One observation to be made is that the grain structure in the WT plane's would not etch with Keller's reagent (which also etches the deformation lines) so WT samples were etched with a dilute nitric acid solution. (See Figure 38 for a comparison of the microstructure revealed by each etchant).

Figure 33: Comparison of cracking of the light and dark inclusions in 2124. Arrows show cracks.
(Rolling direction is vertical)

Cu

Si

Fe

Mg

Mn

Al

Figure 34: Microprobe analysis of a light inclusion
in 2124. (1cm= 5 microns)

Figure 35: Typical microstructure of the 2124 samples
top: high fracture toughness sample bottom: low fracture
toughness sample

Figure 36: Typical microstructure of the 2124 medium fracture toughness sample

Figure 37: Grain structure of 2048 (2124's grain structure was identical) (Slip lines are delineated by arrow)

Figure 38: Grain structure of the WT plane of 2048
after a nitric acid etch.

Discussion

Since the first RMS peak of 2124 in compression and 2048 in both tension and compression occur at approximately yield, much like the RMS peak in pure metals, it would seem reasonable to originally attribute this to a dislocation related process. If this is the case, studies using other methods sensitive to a change in mobile dislocation density should support the assumption. Carpenter working with 7075-T6 aluminum has in fact supported this observation with damping studies. (11)

Since the first AE peak is dislocation related, the AE output could be affected by orientation in any manner which would effect the pinning process. This includes effects due to variation of grain shape and size, to precipitate spacing and density, and to cold work..

Figure 17 shows the effect of orientation on the AE response of 2048 aluminum in compression. Figure 37 shows the grain structure of each plane of 2048 after etching with Keller's reagent. Notice that while the etchant reveals the grain structure in the LW plane, it leaves grain boundaries virtually unetched in the WT plane. Since the long transverse specimen would have the LW plane perpendicular to its loading axis and the longitudinal sample has the WT plane perpendicular to it, further investigation is warranted.

In the case of the LW plane, deformation markings related to slip lines usually occur at high angles to both the grain boundary and the sample rolling direction. Since Keller's reagent will etch both of these features, grain boundaries become visible. Notice that when the grain boundary nearly parallels the deformation lines, the grain boundary is not usually etched. In contrast, deformation lines which are observed in the WT plane are usually nearly parallel to the sample

rolling direction and to most of the long grain boundaries. Most of the grain boundaries are not etched, probably due to the fact that the deformation lines are nearly parallel to them. (Note that an HNO_3 etch, which is not as sensitive to deformation lines, easily etches the grain boundaries).

If it is assumed that the deformation traces are due to rolling and that the slip traces seen in the WT plane micrographs are indicative of the deformation lines within the plane, then rolling must have caused dislocation motion mainly along the set of planes shown in Figure 39.

When the samples are compressed along the long transverse direction, slip planes on which deformation had taken place due to cold work would be parallel to the loading axis and would feel no resolved shear stress. In order for deformation to occur, dislocations would be generated on new planes. The AE response should be much like that of an annealed material.

On the other hand, a loading in the longitudinal direction would cause a resolved shear stress to be felt by dislocations on the cold worked planes. Dislocation motion would then occur on both previously deformed and newly activated planes. The overall AE response should be decreased due to the decrease in mobile density. This in fact is the case.

Figure 18 shows that in the case of tension, only a slight difference in the AE response with orientation is seen when filters are set at 100-300Khz. Since the cold worked dislocations were introduced during compression, a shear stress caused by a tensile force would cause the dislocations to move away from their pileups. During this process pinning should occur and the AE response should vary with orientation

Figure 39: Rolling has caused slip along a set of planes found by rotating plane 1 about the T axis.

since the shear would disappear when samples were loaded in the long transverse direction. Theory and experiment then seem to be in disagreement. Further investigation proves that this is not the case since frequency effects have not been considered.

Figures 20 and 21 show the effect of changing the band pass filter frequency band to 15-300Khz. In the case of the long transverse specimen, an increase in width of the frequency band causes a drastic increase in the AE response. The same increase in band width causes little or no effect in the longitudinal specimen. Since this is the case, a comparison of AE responses of long transverse and longitudinal orientations within the 15-300Khz range of frequency shows the expected orientation differences.

A possible cause of the difference in frequency dependence of the planes deals with the length of dislocation line which is moving. In the long transverse specimen in tension the dislocations generated during the cold working process are moving. In the longitudinal sample new dislocations must be generated. Since the newly generated dislocations see the dislocations introduced during cold work as barriers, the average dislocation length should be less in the longitudinal sample. Also, since longer dislocations have a lower frequency of vibration, and since according to Fleischmann, et. al., (34), this would relate to a decrease in AE frequency, the lower frequency from the long transverse sample would be expected.

In tension (Figure 22), the 2048 sample shows two RMS peaks yet the 2124 curve has only one. In compression (Figures 23 and 24) both materials have two peaks. Hamstad, et. al. (30) has observed that in the case of 2124, after normalization for strain rate and volume effects, the emission peak in tension is approximately 100 times greater than

that in compression. This suggests that these peaks are not caused by dislocation related mechanisms since the deformation process should be virtually the same in both tension and compression.

Graham, et. al. (10) have suggested that the mechanism causing the emission in aluminum alloys is the cracking of second phase inclusions within the matrix. Bianchetti, et. al. (9) have supported this in a paper which relates burst emission in 7075-T651 to the fracture of large intermetallic inclusions. Van Stone (29) has further supported this by observing that the second phase particles begin to fail by cleavage at about 0.7% strain and 50% have failed by 5% strain. The RMS peak usually occurs in the 1 to 5% strain regime. Other authors such as Carpenter and Higgins (11), Hamstad, Bianchetti, and Mukherjee (30) and Mirabile (28) have accepted this theory.

The single AE peak shown in Figure 22 for 2124 in tension is a peak caused by particle fracture. (The peak due to dislocation motion is overshadowed by the peak due to particle fracture). The second peak for the 2048 sample is also related to particle fracture. When the fracture mechanism governs the AE response, RMS peak amplitudes are controlled by two factors: the frequency of cracking of inclusions and the amount of energy released during each cracking event.

In the case of the 2048 peak which has magnitude 5.3 times less than the medium fracture toughness peak and 4 times less than the low fracture toughness peak, the magnitude decrease is expected since both a decrease in the incidence of cracked particles in 2048 and a decrease of energy release due to a decrease in particle size is observed. The decrease in inclusion size also accounts for the fact that the 2048 peak occurs at 3.5% strain, a larger strain value than either the medium

or low fracture toughness peaks which occur at approximately 1.3 and 1.8% respectively, since Hahn, et. al. (24) has observed that larger inclusions crack at lower strains. Also, residual stresses felt by the inclusions affect the position of the peak (35).

In compression, (Figure 24), the 2048 second peak is again lower than the 2124 peaks. This is probably due to the decrease in the incidence of cracking and size of the inclusions as was mentioned previously.

The 2124 second peaks decrease in magnitude in order of increasing fracture toughness during compression testing (Figure 22). However, in tension (Figure 22) the low fracture toughness sample yields less AE than the medium fracture toughness sample. Hamstad, et. al., in a previous article (30), has shown that the high fracture toughness peak in tension lies between that of the low and medium samples. He and his coworkers have related the changes in peak ordering to differences in the direction of tensile stresses felt by the inclusions. Although the exact cause of this orientation effect was not identified during the course of this investigation, the observation that inclusions are slightly elongated in the rolling direction support some type of orientation effect.

In both tension and compression, the medium fracture toughness AE peak exceeds that of the high fracture toughness sample. This is probably caused by the differences in inclusion density observed. That is, since inclusions are about the same size in both samples, we would expect a lower RMS peak from the sample with fewer inclusions.

It would seem that from the previous discussion that the plate fracture toughness is dependent on both inclusion size and density.

Kasteele and Brock (31) working with Al alloys have in fact written an equation relating the fracture toughness to these factors. That is:

$$K_{IC} = C \lambda^{1/2} f^{-1/6m}$$

K_{IC} is the plane strain fracture toughness, λ is the spacing between inclusions, f is the volume fraction of inclusions, and C and m are constants.

In the case of 2000 series aluminum, a decrease in the copper content of the alloy seems to reduce the size of the inclusion and thus increases the fracture toughness. Silicon reduction may have the same effect. A reduction in iron content, on the other hand, seems to reduce the number of inclusions rather than their size, but also increases the fracture toughness. AE seems to be sensitive to both of these effects and in fact seems many times to be more sensitive to the overall fracture process than microscopic observation.

V. Summary

AE has potential as both a quality control and a research tool. AE is more sensitive to microstructural characteristics than many other more expensive tests. In the case of "pure" materials, AE offers new insight into mobile dislocation density changes within the material and thus offers insight into a process which was otherwise almost impossible to evaluate. In the case of alloys, AE can be used to detect the presence of large (45-120 μm) inclusions and in many cases could possibly be used to check whether or not a material meets minimum fracture toughness requirements. This would have the advantage of cutting costs of testing as well as assuring a more dependable group of materials which would save in design costs.

NOTICE

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VI. Conclusions

7075-T651 Aluminum Alloy Study

- 1) The large inclusions are 45 to 120 μ in all dimensions and are segregated at different places within the matrix.
- 2) Many untested inclusions (approximately 40%) contain fine cracks.
- 3) Many inclusions within the plastic zone contain both fine and wide cracks.
- 4) All large inclusions within the tensile sample contain numerous wide cracks. Fine cracks are also observed.
- 5) Wide cracks which are observed, are perpendicular to both the sample rolling direction and the major axis of the inclusion.

Pure Metal Study

- 6) AE from pure metals is related to the dislocation unpinning process.
- 7) A theory using a modified Gilman theory for dislocation mobility fits the AE data for Cu, Ni, Al, and Zr well.
- 8) Cold work decreases the AE from pure Cu.

Aluminum Alloy Study

- 9) The first AE peak in Al alloys is dislocation related.
- 10) Fracture toughness and AE are effected by both particle size and density.
- 11) Filtering effects are important.

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