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ON THE CORRELATION OF ACOUSTIC EMISSION GENERATED
DURING UNIFORM BIAXIAL LOADING TO MICROSTRUCTURAL
SOURCES IN 7075-T651 ALUMINUM AND
21Cr-6Ni-9Mn STAINLESS STEEL

Edward Leon
Dr. Mukherjee
University of California, Davis

December 1981

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On the Correlation of Acoustic Emission Generated During
Uniform Biaxial Loading to Microstructural Sources in
7075-T651 Aluminum and 21Cr-6Ni-9Mn Stainless Steel

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Final Report

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Authors: Edward Leon* and
Dr. A. K. Mukherjee
Division of Materials Science
Department of Mechanical Engineering
University of California
Davis, California

Submitted to: Materials Engineering Division
Lawrence Livermore National Laboratory
Livermore, California

*Currently employed by Hewlett-Packard Co., Loveland Instrument
Division; Loveland, Colorado

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ABSTRACT

This paper reports on the effect on acoustic emission (AE) of uniform biaxial loading of a special thin walled tubular specimen designed by Hamstad, Patterson and Mukherjee. In their initial investigation involving this technique they observed that the AE generated during biaxial loading of 7075-T651 aluminum had several anomalous features when compared to tensile generated AE. Of particular interest, the biaxial AE data was of a much higher level in volts RMS and peaked at a lower strain than the uniaxial AE response. To account for this they proposed a particle cracking model based upon the concepts that inclusions with the largest projected surface area perpendicular to the principle axis of applied loading will crack before smaller inclusions, and that the resulting energy released per AE will be proportional to the crack surface area. In later work 21Cr-6Ni-9Mn stainless steel was shown to exhibit similar anomalous AE behavior. For this study an extensive optical microscopic investigation was performed in order to characterize the inclusion contents with respect to parameters considered to have a first order effect on AE, specifically size, shape, density, hardness, and fracture/decohesion behavior. The inclusions in both 7075-T651 and 21-6-9 display the preferred cracking orientation predicted in the Hamstad, et al model and are shown to be associated with the generated AE. However, other factors appear to contribute to the total AE responses. There is evidence that for 7075-T651 subjected to biaxial loading, a grain boundary related mechanism becomes a significant source of AE in the latter stages of strain hardening. Also, in the case for both materials, an interaction between the complex applied load during

biaxial loading appears to effectively "amplify" the level of AE. The magnitude of this "amplification" effect is defined as the difference in volts RMS of inclusion related AE generated during biaxial loading and that produced during uniaxial loading of a through-the-thickness specimen. The results of this study demonstrate the potential for application of the biaxial technique in AE studies of materials such as 21-6-9, which are notorious for low level AE during tensile testing at room temperature.

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INTRODUCTION

Acoustic Emission (AE) has been defined by the Acoustic Emission Working Group (1) as 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, or alternately as the elastic transient waves so generated. AE generated from metals and alloys is usually associated with plastic deformation of the matrix.

The energy level and characteristic signature of the AE response is highly material specific. For certain materials, such as aluminum and its alloys, the AE response is high and as a result, numerous commercial applications of AE monitoring techniques have been realized. Bailey and Pless (2) is just one group that has demonstrated 'on-line' monitoring capabilities. They successfully equipped an aircraft with the appropriate apparatus for detecting incipient failure due to crack initiation and propagation. Hamstad and Mukherjee (3) demonstrated the potential applications of AE monitoring for process control. In that study, samples of commercial 7075-T651 were subjected to tensile loading and monitored for stress vs. strain, AE, and important deformation properties such as fracture toughness. The stress vs. strain data was nearly equivalent for all specimens but the AE varied decidedly and corresponded favorably with the other mechanical properties, although no difference in microstructural characteristics was detected for the specimens. Fracture toughness is known to be a function of the inclusion content in this material (4) as is AE (5).

Other materials with important industrial applications emit extremely low AE when plastically deformed during tensile loading. Certain steels

(6) have been documented. Usually the AE is too near background, low signal-to-noise ratio, to afford proper characterization. For some materials the AE is somewhat above background but not distinct enough. This property has two serious ramifications in the context of the above discussion. Industrial application is somewhat impaired because of reduced opportunity for process control and on-line monitoring for incipient failure. Secondly, scientific efforts to further current understanding of AE methods suffer from the absence of otherwise desirable material systems.

Progress has been made on an AE technique which may facilitate opportunity for study of low tensile AE materials. Hamstad et al (7) designed a special thin walled tubular specimen which, when internally pressurized, imposed uniform biaxial loading. In their initial results from unflawed 7075-T651, they reported high levels of AE when compared to tensile AE data. Other AE anomalies were observed and they proposed an inclusion cracking model to account for these differences. The original motivation of the work was to provide means of subjecting the unflawed matrix to a complex stress state in the same vein as fracture toughness loading of flawed (precracked) matrix. While they demonstrated that there was a potential for detecting the difference between AE generated during normal deformation processes (plastic deformation and inclusion particle fracture) and crack extension, the technique displayed 'amplification' characteristics.

The objective of this paper was to investigate the 'amplification' effect as it relates to the complex stress state imposed on inclusions during biaxial loading. A secondary concern was providing further evidence system noise was not a problem and that the AE was indeed

representative of sources within the deforming gage section. Finally, of a more practical nature, information was desired that would facilitate a general statement about the order of magnitude of 'amplification' one might expect when using this technique.

In pursuance of these objectives, an extensive microstructural examination was performed on 7075-T651 aluminum and 21Cr-6Ni-9Mn stainless steel and then correlated to the AE generated during uniform biaxial loading. The specimens were obtained from M. Hamstad after being tested at Lawrence Livermore Laboratories.

In the next section, the AE testing background will be reviewed, followed by a discussion of the practical considerations which motivated the design of this study.

BACKGROUND

The initial testing and characterization of the biaxial method was done with 7075-T651 specimens machined from commercial bar stock (7). This material was known to emit high levels of AE; its behavior during tensile loading being well documented in the literature (8,9,10). Moreover it was relatively easy to machine and familiar to Hamstad et al as evidenced by the number of their related works (3,5,10-12). The results are illustrated in Figures 1 and 2 for biaxial specimens subjected to biaxial and uniaxial (tensile) loading respectively. The tensile AE response is similar in shape and amplitude to AE results using the usual pin loaded longitudinal tensile specimen (12). The biaxial data is different than both in several important aspects. The overall RMS voltage level is higher and has several bursts superimposed upon the continuous signal. In addition the AE appears to peak (rise, reach a maximum, and taper off) at a much lower level of strain. Furthermore, for cycled loading mode tests, when tensile loading immediately followed biaxial, very low levels of AE were observed.

Hamstad et al (3) propose a particle cracking model to account for their observations, based upon the concept that inclusions have a tendency to crack perpendicular to the principle axis of the applied load. This model will be discussed in more detail later in this paper, but for now let it suffice that the anomalous features in the biaxial AE response are purportedly a consequence of the applied load/specimen-geometry interaction. Hamstad et al theorize that inclusions in 7075-T651, which they generalize as spherical ellipsoids, crack along

their long axis during biaxial loading and their short axis during tensile loading. Hence, because the amount of energy released during cracking is proportional to the crack surface area, the resultant AE would be higher for the biaxial loading condition. They maintain that the AE would peak at a lower strain because the inclusions would crack at lower strain levels. This idea was based upon an observation by Van Stone et al (13) where particles having the largest cross-section perpendicular to the tensile axis crack first at lower applied stress levels while those with smaller cross-sections crack later at higher applied stress levels.

Hamstad et al recognized that even if their theory about particle cracking was valid, the AE data appeared influenced by what they term "applied tensile stresses in two directions". They did not expound upon this further but the following argument lends credence to this concept. The tensile AE from the biaxial specimen, Figure 2, afforded direct comparison to the AE data generated during biaxial loading, Figure 1, because the specimen geometries were identical. The tensile data compares favorably with longitudinal tensile data from other tests (12) for plate material even though the specimen configuration was different, thin walled tubular versus pin loaded tensile; in fact the gage section is nearly 30% larger. James and Carpenter (14) showed this to be an important parameter when considering tensile AE from this material. It appears, however, that other factors in the system offset the effect of the gage volume difference. Now consider the AE data in Figure 3 for a through-the-thickness tensile specimen of the same dimensions as the pin loaded specimens considered from reference (12) above. The level of AE is higher than the longitudinal tensile

data. If the anomalous high AE response for biaxial loading was due only to the long axis of the inclusions perpendicular to the major axis of loading the tensile AE in Figure 3 for the through-the-thickness should be the same as the biaxial because the inclusions are comparably orientated. It is much lower and it is concluded that some other source is responsible, perhaps a "complex state of stress around the inclusions" as Hamstad et al suggest.

Physical constraints of the stock material prohibited tensile loading of through-the-thickness specimens machined in the biaxial configuration so resolution of the question required a different tact. A second material was chosen to study, 21-6-9 stainless steel, which was known to have several types of inclusions (15), but low tensile generated AE at room temperature from Hsu et al (16) on a similar alloy. Bar stock was chosen so that the inclusion segregation characteristics would be similar to 7075 aluminum, i.e., aligned along the bar axis. To insure isotropic grain characteristics, a special heat treatment was used, to be discussed in a later section.

The AE data illustrated in Figures 4 and 5 were generated from biaxial and tensile loading, respectively. The specimens corresponding to AE records #1 and #2 were machined from two batches of air remelt material (to be known as air melt) while numbers 3 and 4 are from two different batches of vacuum melt stock. The RMS levels for all biaxial tests are significantly higher than those for the tensile tests. It is interesting to note that among the biaxial specimens the RMS level of air melt AE records is generally higher than the vacuum melt and radically different for the two specimens. The two vacuum melt AE signatures on the other

hand are very similar. Similar trends appear to apply to the tensile AE data but the signal-to-noise level is so low that general statements are not feasible.

PRACTICAL CONSIDERATIONS

7075 aluminum can be precipitation hardened. It has a sequence of precipitates and secondary phases ranging in size from a few atom clusters to several microns (17). Material treated in the T6 and T651 tempers have received considerable attention in AE studies (8-10); the T651 and T6 are identical except that the former is stretched between 1.5 and 3% after re-solution and quench, prior to aging. Simultaneously monitoring dislocation damping and AE, Higgins and Carpenter (9) were able to correlate the tensile generated AE in 7075-T651 exclusively to particle cracking. Several researchers, on the other hand, have attributed AE in 7075-T6 to both dislocation and particle cracking mechanisms (12, 18).

The primary objective in this part of the study was to evaluate the accuracy of the particle cracking model Hamstad et al proposed to explain the anomalous AE features of the biaxial specimens, Figure 1. To this end the particle cracking orientation was characterized by means of a microstructural investigation. Efforts to correlate AE to the matrix phenomena were based upon a study by Swindlehurst (19). He was seeking to determine the relationship between carbide cracking and AE in a low alloy steel. He subjected tensile specimens to different levels of strain and quantified the number of fractured carbides corresponding to each strain level using optical microscopic and scanning electron microscopic techniques. A subsequent plot of the density of cracked particles as a function of strain compared to the AE indicated that the AE was not coincident with the cracking of the particles, i.e., carbides were cracking at strain levels when no

AE was generated. Swindlehurst concluded carbide cracking was not a significant source of AE in low alloy steel. It was hoped that this study on 7075-T651 would similarly yield a definitive statement, information completely corroborating or contradicting the model of Hamstad et al.

Using the Scanning Electron Microscope (SEM), the fracture surface of biaxial specimens were also examined to provide more information regarding the participation of the inclusions in the deformation and fracture processes.

The principal sources of emissions from industrial steels appear to be the following (20):

- (a) The sudden cooperative motion of groups or "packets" of dislocations.
- (b) The fracture of brittle or semi-brittle second phase particles.
- (c) Crack initiation and/or propagation, where this differs from (b).
- (d) Military phase transformations.

For 21-6-9 subjected to the loading conditions under consideration only (b), the fracture of brittle or semi-brittle second phase particles, appears to be a viable source. Consider the observations of these other workers. AE due to dislocation activity, (a), was negligible for a similar material subjected to tensile loading at room temperature (16). The specimen configurations are designed to be unflawed to minimize the influence of source (c), crack propagation and/or extension. 21-6-9 is an austenitic stainless steel; a group of materials not usually associated with (d), military phase transformations (21). In addition per the results of Swindlehurst above, and others (22), carbide cracking is also probably not a source.

This part of the study was directed at basically one thing, correlation of the AE to aspects of the inclusion contents of the four stock materials. This involved initially classifying inclusion types on the basis of size, shape, color and composition. Once this was completed, inclusion properties were tabulated which were thought to have a first order effect on AE, which in addition to size and shape (projected surface area) included hardness, density, and cracking/decohesion behavior. An optical microscope was used to perform the microstructural examinations.

MATERIAL AND METHODS

Background Experimental Considerations

All mechanical testing was performed at L.L.L. under the auspices of Dr. Marvin Hamstad. Certain experimental parameters are considered pertinent to this paper and are presented. For a more detailed account of the testing procedure including the various measures taken to ensure reliable data, the reader is encouraged to peruse reference (7).

Biaxial loads were applied by internal pressurization of special thin walled tubular specimens. The setup for this configuration is illustrated in figure 6c while the specimen geometry is shown in figure 6b. A threaded plug is inserted in the open end and oil impressed into the cavity. By means of feedback control the hoop strain rate, measured with clip strain gages, is kept relatively constant. This rate was 0.2%/min for all tests. The AE signal was amplified to a gain of 86 dB and 88 dB for the 7075-T651 and 21-6-9 specimens respectively.

To facilitate direct comparison to the biaxial AE data for 7075-T651 in figure 1, thin walled specimens of dimension identical to those used in the biaxial. Tests were pulled in tension and the resulting AE, figures 2 and 5, amplified to comparable gain levels. The tests were performed using a standard tensile machine, see figure 6a. The tensile data for 21-6-9 was obtained by the application of a tensile load on specimens machined in the conventional tensile shape, figure 7, and subsequent amplification to 90 dB gain. The crosshead velocity was 0.02 mm/sec for both tensile specimen types although the effective strain rate was 2.2%/min for the 7075 compared to approximately 5%/min

for the 21-6-9.

All the specimens discussed to this point were machined from bar stock in a longitudinal orientation, that is the long specimen axis coincides with the bar axis of the material. Other 7075-T651 specimens were machined from plate stock; longitudinal and through-the-thickness (perpendicular to the rolling plane). Both were dimensioned according to figure 7 and loaded to failure. The AE generated during deformation of the through-the-thickness specimen is depicted in figure 3.

The 7075-T651 aluminum bar stock was 51 mm in diameter with a standard T651 temper, that being a stretch of 1-3% after resolution and quench, prior to aging. The composition of this material is shown in table 1.

While all 7075 specimens were machined from essentially the same stock, the 21-6-9 samples were machined from material from four date codes. The composition for each is given in table 2. Two 2.5 inch diameter bars of air melt material and two bars of 2 inch diameter vacuum melt were selected. After machining, these samples were annealed in order to remove the effect of remenant plastic work. Each specimen was heated to and maintained at 1010°C (1850°F) for four hours in vacuum, then while still under vacuum allowed to furnace cool relatively rapidly (approximately 1 minute) to 650°C. After back filling the furnace with argon, cooling was allowed to continue until ambient temperature was reached.

Optical Microscopy

Polishing techniques for revealing the 'true' microstructure of

metals for examination with a light microscope are discussed at length in several publications, the ASM (23) and ASTM (24) handbooks are notable examples. For the most part the procedures discussed in these works were followed in this study. The tested specimens were sectioned to a manageable size, typically 1cm x 1cm to afford insertion in the SEM specimen chamber, and mounted in molding plastic. The roughest irregularities were eliminated by graduated sanding on 240 through 600 grit paper. Rough polishing was accomplished on a polishing wheel at low speed using 15, 9, and 6 μm diamond paste and final polishing using the smaller grit diamond, 1 and $\frac{1}{2}$ μm , and 0.3 and 0.05 μm alumina, also at low speed. Micrographs were obtained using a Bausch and Lomb Bench Metallograph System with photographic attachment.

Polishing procedures varied depending upon material considerations and whether the desired information related to grain or inclusion structure. Some aspects of these procedures are worth noting because although they are inherently unique for the material and applied loading condition, the concepts are transferable to similar material systems.

Several characteristics of the 7075-T651 matrix made it a difficult material to polish properly. There were a large number of hard inclusions dispersed in a relatively soft matrix. Inherently low cohesion forces promoted inclusion fallout during polishing resulting in a loss of information and poor surface finish. This condition was accentuated for material previously exposed to high levels of plastic deformation whereby massive particle fracture effectively reduced the cohesive force. To further complicate matters the detail of the remaining inclusions was often obscured by matrix material which 'flowed' during polishing.

Surface preparation for grain structure inspection was facilitated

by polishing with 15 μm diamond grit succeeded by 9, 6, 1, and $\frac{1}{4}$ μm polish. Ample grain structure was revealed by light etching with Kellers reagent. This range of grit sizes was too large to allow proper polishing for inclusion inspection because of excessive fallout. Accordingly the rough polish stage was limited to a light 6 μm polish followed by 1 and $\frac{1}{4}$ μm diamond polish. The final two grit sizes had to be repeated several times before the distorted and flowed material was removed. The final polish was especially critical for revealing the cracking characteristics of the inclusions in the pressurized specimens.

For 21-6-9, the scratches from 600 grit paper were rather easily removed by polishing with a 0.3 μm alumina and water mixture. The final polish usually only required two or three minutes of 0.05 alumina and water. The cohesive forces between the various inclusion types and the matrix are high and as a result very few are pulled out. This was verified by constant revisiting of a selected area during polishing to verify retention. The inclusion detail was excellent, a consequence of little matrix 'flowing' and the high contrast between inclusion and matrix.

Variations in the passive layer thickness made etching a delicate proposition. Chemical etchants, Viellas and Marbles (25), were not satisfactory because of too little and too much aggressiveness respectively. Efforts to polish electrolytically with Kallings and Frys (25) reagents proved equally ineffective. The etchant composition used was suggested by Arieli (26), and consisted of 25 g FeCl_3 , 40 cc HNO_3 , 15 cc H_2SO_4 , 240 cc HCl , 40 cc CH_3COOH , and 100 cc H_2O . The etchant was applied with a cotton swab and cleaned with methonal and water. This procedure, final polish and etch, was repeated several times until the

disturbed material was removed.

Scanning Electron Microscopy (SEM)

The procedure for specimen preparation was straightforward. After being sectioned, each sample was degreased by a methanol flush and then glued to an aluminum mounting stub using the adhesive properties of Silverprint conductive print. A Cambridge Stereoscan scanning electron microscope (SEM) was used.

Variation in topography is the principle source of contrast in the secondary collection mode of the SEM, but can be a problem for fracture surfaces if the gray scale of the collection and display apparatus is unable to accommodate the magnitude of the contrast. In the case of 7075-T651 for common accelerating voltages of 10-20 Kv, peaks and ridges became over-saturated while no detail was evident in the depressions. Under certain extreme conditions oversaturation of the phosphorous in the CRT caused streaking on micrographs until de-excitation of the phosphorous. One method considered to rectify this situation was gold sputtering. Deposition of the high Z material drastically reduced the contrast variation but resolution suffered due to surface artifacts of gold. The best results were obtained by lowering the accelerating voltage while simultaneously varying condensor lens current, stage tilt, scan time, and final aperture size to offset the attendant degradation of resolution. For low to medium magnifications this corresponded to an accelerating voltage of 1-5 eV and increased lens current, stage tilt equal to 35-45, reduced scan time, and smaller aperture size. For larger magnifications an accelerating voltage of 5-10 eV was required

for adequate resolution.

Energy Dispersive X-ray Analysis (EDXA) with SEM and Electron Microprobe

Inclusion composition was determined by analysis of the characteristic x-rays emitted by the occurring elements when excited by the impingement of a high energy electron beam. This was facilitated using the appropriate EDXA apparatus fitted to an electron microprobe and a SEM. The electron microprobe with its high energy beam and large spot size was appropriate for compositional dot 'maps' for which the beam was rastered across a $50\text{ }\mu\text{m} \times 50\text{ }\mu\text{m}$ area registering a dot when the number of energy counts of the desired x-rays were above a specified threshold level. For more delicate work such as element determination of deposits within inclusions, the SEM was ideally suited. Standards were not available so the quantitative compositional data was only approximate.

The procedure for specimen preparation was identical to that for the optical specimens. To avoid biasing of the EDXA data, each specimen was lightly etched immediately prior to insertion into the SEM specimen chamber. Before insertion, however, each specimen was removed from its plastic mold. This was achieved by carefully sawing around the edges and gently prying it loose.

RESULTS

7075-T651

The general characteristics of the grain structure and inclusion distribution for bar stock 7075-T651 are depicted in figures 8a and 8b, respectively. Similar to plate material in the sense that the grains are elongated with axis parallel to the rolling or bar axis, the grain structure for bar stock material nonetheless differs in that the grains are not flattened but somewhat football shaped. It is evident from figure 8b that this material has high inclusion density, estimated at 10^{-7} - $10^{-8}/\text{cm}^3$ by Hamstad, et al (5), that commonly segregate into stringer formations.

Two types of inclusions were observed in the reduced section of the thin walled specimens. The first type, hereafter to be referred to as dark inclusions were comprised of nearly equal proportions of Mg and Si. This is pictorially displayed in figure 9 by elemental composition maps produced using the electron microprobe. An area in the vicinity of the inclusion lighter than the background, indicates a greater amount of the considered element in the inclusion than the matrix, while a dark area indicates just the opposite. The dark particles were typically equiaxed ranging in size from 1 to 5 μm , and often times coincided with grain boundaries. The second type, light inclusions were composed of Fe and Cu with a trace amount of Cr. The existence of Cr was detected using EDXA in an SEM. The light inclusions were generally larger than the dark, 5-20 μm in the longest dimension and elongated with the long axis parallel to the bar axis. Other research-

ers have observed inclusion of comparable shape, size, and composition (8).

Both light and dark inclusions are shown in figures 13a, b, and 14a, and b. These micrographs are of material which had been subjected to biaxial (internal pressurization) and tensile loading and display a preferred cracking orientation. It is evident that the particles crack perpendicular to the major axis of the applied load. In the case of biaxial loading, figure 13, where the hoop stress was twice that of the axial, the light colored inclusions have cracks parallel to the bar axis. When the applied load was exclusively axial, figure 14, the light inclusions displayed cracking perpendicular to the bar axis. The conditions are expressed schematically in figures 13c and 14c for each case. Note that the arrows in the diagrams represent applied loads on the specimen; the actual stresses on the inclusions are much more complex. In the uniaxial case for instance it is widely accepted that Poisson compressive stresses react the particle perpendicular to the tensile axis (27).

Other aspects of figures 13 and 14 deserve mention. The cracks for the particles in figure 14 are much wider than those in the particles of figure 13. This can probably be attributed to the much higher level of strain for the tensile test, 10% vs. 2.5%. Also, cracking is only evident for the light inclusions. There are several possible reasons for this. First, as previously mentioned, the dark inclusions are typically smaller than the light. Van Stone et al (13) reported they observed large inclusions crack first at low applied strains and smaller inclusions ultimately crack at much higher strain levels. For the case of the dark inclusions, macroscopic failure may precede strain levels sufficient to induce cracking. Secondly, there would be little contrast

between an inclusion and a crack it might have. It is plausible that cracks exist that cannot be resolved optically. A few particles were examined using the SEM and the dark field mode of the Bausch and Lomb but these efforts proved inconclusive. Finally, if dark particles had cracked, it is possible that in their weakened state they were pulled out of the matrix during polishing.

The percentage of the cracked light particles vs. strain for four data sets is plotted in figure 11. Cracks detected in 2-D optical micrographs were verified in several cases using the SEM. The data sample was composed of approximately 50 micrographs of each strain, each a 2-D field of $.01 \text{ mm}^2$ ($100 \mu\text{m} \times 100 \mu\text{m}$), and averaging five inclusions per field totaled approximately 250 inclusions. More than one crack was allowed per inclusion and only inclusions at $5 \mu\text{m} \times 5 \mu\text{m}$ and larger were considered because of resolution problems.

The fracture surface of the longitudinal plate tensile sample is shown in figures 14 and 15. Depicted is the ductile tear morphology of a 'quasiductile' material. Although the matrix is very ductile, only small dimples are observed, most likely nucleated at large aging precipitates (27). Heavy alloying and the resultant high inclusion density render 7075-T651 much less ductile than pure or low alloy aluminum(28).

The reduction in ductility was more pronounced for specimens pulled in the through-the-thickness direction as evidenced in figures 16 and 17. The morphology is described as a succession of 'steps'. These 'steps' can be reduced semantically to shear walls and plateaus as portrayed in figure 17. It is probable from the conspicuous locations of inclusions on the fracture surface, see figures 17b and c, that they play a major role in the fracture process for this orientation.

The morphology of a specimen loaded to failure by internal pressurization is characterized by the sequence of micrographs in figure 18. At low magnifications the surface has a stepped appearance, but at intermediate magnification the morphology is much more complex and in fact is a characteristic of several modes. In 18c and d we see the cup and cone characteristic of a ductile material with uniform particles. Figures 18e and f display evidence of grain boundaries participating in the crack path. Approximately 10% of the fracture surface could be characterized as ductile cup and cone, 10% intergranular, while the largest portion was undescrivable in terms of classic fracture terminology, i.e., mixed mode. Each group appeared to be randomly dispersed.

The micrographs in figure 19 represent a different approach to analyzing fracture characteristics. The fracture path in b for the longitudinal tensile specimen is exclusively transgranular. For the through-the-thickness tensile specimen, figure 19a, the fracture path almost completely coincides with grain boundaries occasionally transverseing the matrix to link with the next grain boundaries crack. The effect is not as pronounced for the biaxial specimen figure 19c although there is evidence of some grain boundary participation.

For specimens where the principle axis of loading is perpendicular to the rolling plane or bar axis a probalbe failure scenario proceeds as discussed in figure 20 (29).

- i Inclusions fracture at low strain levels
- ii The crack propagates along the path of least resistance which is grain boundaries and grain boundary particles
- iii Cracks link via transgranular fracture
- iv Propagation continues until critical crack size is reached

21-6-9

The grain structure and inclusion distribution for air melt and vacuum melt 21-6-9 are illustrated in Figure 21 and 22. Unlike the elongated grain structure observed in 7075-T651, the grains are equiaxed for both treatments with a large percentage having annealing twins. The unetched matrix for both treatments appears relatively 'clean' when compared to 7075. The stringers tend to align into stringers along the bar axis.

Four distinct inclusion types were observed in the 21-6-9 biaxial specimens where classification was based upon size, shape, color, and composition. The chart in Figure 23 provides a summary of the characteristics of each type in the context of their distribution in each specimen. Three types were identified in the air melt specimens; only two for the vacuum melt. Examples of each type are shown in Figures 25-28. Type A is an oxide with Mn-27Cr-17Ti. Typically it was spherical, 2.5-10 μm in diameter, and generally segregated in stringers such as the one pictured in 26a, although a few large solitaires were also observed. Type B, 50Mn-50S, is a commonly observed inclusion in steels (30). For this condition it usually segregated as elliptical spheroid solitaires, Figure 25a, although it was occasionally observed as a scavenging particle on larger inclusions, Figure 28a. There are actually two distinct compounds in type C, large nitrides of composition 40Mg-20Cr-15Ti linked by 50Mg-50Si. The large amount of scooped out material evident in Figures 27 and 28 suggest that the Mg Si segregation is relatively soft with respect to the nitrides. Inclusion type D, 50Mg-50Si, was only observed in the vacuum melt 21-6-9 stock.

Indicative of the inclusions shown in Figure 29, they were generally shapeless, aligned in stringers along the bar axis.

In addition to size and shape as already discussed, other inclusion characteristics considered important for a discussion on AE are presented in Figure 23. These include density, hardness, and cracking/decohesion characteristics.

DISCUSSION

7075-T651

The light particles in 7075-T651 fracture in a manner consistent with the model proposed by Hamstad et al (7). This is clearly in evidence in figures 13 and 14, a and b, in which the preferred cracking orientation is perpendicular to the principle loading axis. Based upon this knowledge a statement can be made concerning a question Bunch (31) had while investigating the cracking behavior of Cr rich inclusions in 7075-T651. It is apparent now that the cracking he observed perpendicular to the rolling, principle stress, and long inclusion axis was because of 'stress variation in the matrix' and not 'specimen (inclusion) geometry' as he tentatively surmised.

This evidence of cracking is consistent with reports by Graham (32) and others that the fracture of intermetallic particles is a source of AE in 7075-T651. Whether the relationship between cracking and AE response proceeds according to the scenario envisioned by Hamstad et al (7) is something that warrants further scrutiny. As you recall, in that study they reported on anomalous characteristics of the AE record for 7075-T651 subjected to internal pressurization. First, the RMS energy level of the quasicontinuous AE response was consistently higher than the comparable record from a thin walled specimen loaded in a uniaxial manner with several discrete bursts very much higher. Also, the AE was observed to peak at a much lower level of strain. Finally, for alternating loading mode cycle tests in which tensile followed biaxial, the AE appeared to be lower. Hamstad et al proposed the particle

cracking theory to explain these observations. They felt that the high AE RMS energy level was a direct consequence of an applied load/inclusion geometry interaction. They maintained that because the projected fracture surface area of the inclusions (considered spherical ellipsoids) would be larger for biaxial loading the resultant energy release would be higher. They suggest the early AE peak was also a consequence of inclusion geometry. Van Stone et al (13) reported that inclusions with the largest cross section perpendicular to the tensile axis crack first at low applied strain levels while those with smaller cross-sections crack later at higher strain levels. Finally, during the biaxial phase of the cyclic loading they surmise that a high proportion of the particles crack and consequently the AE response is effectively muted in subsequent tensile tests. The evidence of cracking in figure 13a and b, corroborate the basic premise of this theory: particles do indeed crack perpendicular to the principle axis of loading.

The data plotted in figure 11 lends direct support to the concept of inclusions as AE sources. No cracking was observed in untested material (this was taken from unloaded grip section). After 0.5%, 30-50% of the inclusions were cracked. The specimen loaded to 0.8% hoop strain appeared to have a slightly higher percentage of cracked particles 50-60%, while the specimen loaded to failure, approximately 2.5% hoop strain, had 60-70% cracked. This indicated that 35-50% crack in the pre yield and early stages of yield, another 10-15 % crack in the 0.3% strain corresponding to the AE peak and a final 10-15% cracked during the remainder of the test. The first three data points provide strong evidence that the particles are responsible for the AE during the lower levels of strain. The last data point is not inconsistent

with the observed AE but does not directly coincide. It appears that 60% of the particles crack during the final two thirds. From figure 2 it is evident that although the AE is decreasing in the strain hardening region, it is still considerably above the noise level. This implies that another source or sources may contribute to the AE response during the higher strain levels of 7075-T651 subjected to biaxial loading.

The possible sources are many. A dislocation mechanism may begin to dominate. Schmitt-Thomas et al (33), Wells (18), and Carpenter et al (9) all reported on a double peak in 7075-T6 and a similar alloy, and attribute the first to dislocation activity. Only a remote possibility exists for a couple of reasons. First the AE levels due to dislocation motion are only 1/10 of that observed and are usually most prevalent near yield before appearing to decrease altogether. Secondly the AE in figure 1 is quasicontinuous; not indicative of dislocation related AE which is very smooth because of the large source size. A more likely source is fracture or decohesion of the dark particles. These small particles would have a tendency to crack only after large strains, per Van Stone et al (13). Unfortunately, as mentioned in the results the morphology was difficult to characterize optically. It was noticed, however, that the dark particles became increasingly more difficult to retain in the matrix during polishing for specimens subjected to higher total strain levels, .05%-2.5%. This could be construed as evidence that particles were cracking or decohering at the high strain levels and falling out during polishing because of a decrease in the cohesion force between particle and matrix. Jaffrey (34) uses a technique based upon this concept to determine the incidence of

cracking or decohesion of hard to retain inclusions in which he compares the retained inclusion density of stressed material to unstressed material.

Another plausible source is crack extension along grain boundaries. For the biaxial loading condition approximately 10% of the fracture surface morphology was characterized as intergranular. Thus at some point in the deformation sequence crack extension proceeded along grain boundaries. It is unlikely that this occurred after fast fracture commenced as the front travels at slightly less than the speed of sound in this material system (35) much too quickly to allow the observed coherent grain boundary separation. Therefore, it must be assumed the grains were mechanically decoupled prior to the onset of fast fracture, although the exact instant of separation is unknown.

Lindly, et al (36) do not consider crack extension per se a significant source of AE in this material. They suggest this explains the observations of Harris and Dunnegan (37) during fatigue testing, where AE bursts were not observed every cycle, yet an electron microscopic investigation revealed evidence of a crack serration for every cycle. In a similar manner perhaps, the AE response for 7075-T651 may be relatively low for crack propagation along grain boundaries while exhibiting bursts when the crack extension process induces decohesion or fracture of the grain boundary particles.

21-6-9

The inclusion types classified in this study are quite common for this alloy (15) and have been observed in other steels. In fact, the hardness values assigned in the chart in figure 23 were culled from data in the Handbook of Stainless Steels (21). It is apparent that the specific inclusion contents of the vacuum melt specimens varied significantly from the two air melt stocks. This can be explained in the context of the processing procedures used in the fabrication of the two melts. The following discussion is based upon a conversation with a representative of Armco (15). In the initial step for both processes materials in the proper proportion are placed into a furnace and smelted in a nitrogen atmosphere at atmospheric pressure. To remove the gross inhomogeneities the molten metal is then poured through filter beds. The billet material at this stage is air melt 21-6-9. For the vacuum melt, or remelt, designation this material then receives an additional processing step designed to refine the inclusion content. The billet is placed in a chamber, a vacuum drawn, and the material is remelted and cleaned using zone refining techniques.

In general, the biaxial AE data compares favorably with these inclusion characteristics. In comparing the AE responses in figure 4 it is observed that the records are quite different for the two air melt specimens. By contrast, the AE records for the two vacuum melt specimens are nearly identical albeit different from both of the air melt responses. Thus, deformation of the vacuum melt specimens shown to have similar inclusion contents resulted in similar acoustic emission. On the other hand, the two batches of air melt stock, products of a less

controlled process, had more diverse inclusion contents and correspondingly different AE. It is also apparent that the AE response of the air melt specimens is generally higher and more discontinuous than the vacuum melt response. This is not surprising considering the contrast in inclusion makeup: silicates and sulphide in the vacuum melt material compared to silicates, sulphides, nitrides and oxides in the more exotic air melt matrix.

Determination of the actual inclusion fracture/decohesion mechanism (s) responsible for the AE is not as straight forward. The vacuum melt AE is probably associated with inclusion type D, a Mg-Si compound, as it was the only prevalent inclusion segregation. Evidence of fracture or decohesion was lacking, however. The air melt case is much more complicated. An attempt was made to decipher the signature differences of the air melt specimens in terms of inclusion content based upon the assumption that distinct mechanisms acted as sources for each AE feature (high or level continuous or burst type) and that each mechanism was associated with a different inclusion type within the matrix. This concept is not unreasonable in light of the results of Bianchetti, et al (5), for the case of 7075-T6 subjected to tensile loading. A few specimens exhibited AE responses which displayed several high level bursts superimposed upon the familiar quasicontinuous record. They surmised that each burst corresponded to fracture of a large Cr rich particle; an inclusion type not commonly observed in this material. As shown in this and several previous studies, the relatively smooth quasicontinuous part of the response is associated with fracture of small, plentiful Fe and Si rich particles.

To return to the main point, it was felt that the high level burst type AE in records No. 1 and No. 2, figure 1, was the result of the fracture of a few very large, brittle inclusions of a single type present in both air melt systems. The continuous type AE in No. 1 was thought to be related to cracking of a large number of a large or medium sized inclusion type that would be present in No. 1 but not in No. 2. Finally, for the low level continuous AE in No. 2, the cracking of a large number of a small inclusion type was considered a possible candidate. The last type could be present in No. 1 also, the AE being masked by the higher level continuous AE. Unfortunately, as can be seen in figure 23, the observed inclusion types and respective characteristics do not fit this scenario. There was one inclusion type in specimen No. 1 that was not observed in No. 2, that being type C, and although type C inclusions were generally large, only a small number of them were present. This last condition conflicts with the generic nature of quasi-continuous AE whereby a response is generated by a large number of events. Thus, it is not obvious how to account for the fact that high continuous AE was generated by No. 1, but not No. 2. Considering the unknowns inherent in the air melt fabrication procedures, further hypotheses as to which inclusion might be responsible for which AE feature would be pure speculation.

Another important feature of the acoustic emission is the large difference in magnitude of AE measured as volts RMS between the biaxial records in figure 4 and the tensile records in figure 5. This variation appears to be a function of the biaxial loading mode and the specimen-geometry/applied-load interaction. This will be discussed in more detail in the final remarks and comments.

SUMMARY

For the most part the particle cracking behavior and associated AE observed in this study is consistent with the model proposed by Hamstad, et al (7). The inclusions displayed a preferred cracking orientation perpendicular to the principle axis of applied loading. This is evident in figures 12 and 13 for 7075-T651 and figures 25-28 and 30 for 21-6-9. Also consistent with the model is strong evidence to suggest that particle fracture is intimately related to the AE generated in these materials. For 7075-T651 a plot of percentage of particles cracked vs. applied strain level shows a direct correlation to the AE response during biaxial loading. In other words, when particles were cracking, AE was observed. For 21-6-9 the AE response and inclusion distribution appeared to be function of the melt. The relatively "clean" matrix of the two vacuum melt specimens generated consistent AE when deformed while the AE from the air melt stock, products of a less refined process, was drastically different for the two specimens.

One of the premises in the Hamstad, et al model is that the amount of energy released via AE is a function of the mean crack surface area of the cracking particle. Thus, in the case of 7075-T651, when the inclusions crack along their long axis during biaxial loading, the resulting AE would conceivably be higher than AE generated during tensile loading when particle cracking would be predominately along the short axis. To a certain extent the results of this study corroborate this concept. The level of AE measured as volts RMS was higher for biaxial loading than tensile loading. This is clear when comparing figures 1 and 2, and 4 and 5. Upon closer inspection the magnitude of the higher

AE response during biaxial loading appears to be more than a function of a larger mean crack surface area of the inclusions. The AE generated during tensile loading of a through-the-thickness specimen, figure 3, was higher than tensile AE data from a longitudinal specimen, figure 2, but not nearly as high as the biaxial response, figure 1. The principle axis of loading for both the through-the-thickness tensile and the thin-walled specimen subjected to biaxial loading is perpendicular to the long axis of the inclusions. As noted in the Background section of this paper, the effect of gage volume difference on AE in this case does not appear to be significant. Thus, the AE should have been the same if crack surface area were the the only influence on energy release.

This argument is not entirely valid for, among other reasons, the unknown influence of other sources on the AE response. As shown in this study there appears to be another active source, perhaps as was suggested, crack extension along grain boundaries or fracture of grain boundary particles induced thereof. To reduce the chance of other active AE sources being present, the 21-6-9 specimens were subjected to the special annealing treatments outlined in the Materials and Methods. When testing began, there was a reasonable degree of confidence that the AE would be generated by particle fracture and not be affected by anisotropy with regard to the inclusion distribution, although perhaps not entirely. There was no evidence of crack extension from particle to particle within inclusion stringers, a known source of emission, so it appears the particles behaved independent of one another. The inclusions were somewhat elongated but not to the same extent as the light particles in 7075-T651; i.e., lower length to width ratio.

Furthermore, there was no difference in severity of cracking among the inclusions in the tensile and biaxial specimens. This latter point is illustrated if the inclusion cracking figure 30, for an applied tensile load is compared to the inclusion in figure 28, for an applied biaxial load.

It has been shown that a particle related AE response appears to be effectively amplified during uniform biaxial loading of unflawed specimens. Hamstad, et al (7) suggest that this is the result of the effect of the complex stress state surrounding the inclusions caused by the applied tensile stress in two directions during biaxial loading. The magnitude of this effect is the difference in the RMS level of AE generated during tensile loading of the through-the-thickness specimen and AE generated during biaxial loading of the thin-walled tubular specimens.

The results of this study are summarized in the following statements:

7075-T651

1. AE generated from plastic deformation of 7075-T651 is due to inclusion cracking and the level of AE measured in volts RMS is related to the mean crack surface area.
2. For biaxial loading another AE source appears to be active during the latter stages of strain hardening. There is some evidence that it is related to a grain boundary mechanism.

21-6-9

3. The AE response is much higher for 21-6-9 subjected to uniform biaxial loading than for tensile loading. The AE generated during tensile loading of longitudinal specimens is too low

to characterize properly.

4. The AE response during uniform biaxial loading appears to be a function of the inclusion content which was different for the two air melt specimens but the same for the two vacuum specimens. The inclusion content is characteristic of the fabrication processes for the different melts. The inclusion fracture/decohesion mechanism (s) responsible for the AE is not immediately obvious.

General

5. The AE records generated during biaxial loading appear to be amplified somewhat when compared to AE records generated during tensile deformation of comparable orientation. For 7075-T651 an increase in AE of 25% was realized for AE measured in volts RMS.
6. The biaxial loading technique has the capability of generating reproducible AE records that are indeed representative of microstructural deformation events. Noise does not seem to be a significant problem.

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Figure 1: Typical test results for an internal pressurization test to failure: (a) axial strain, hoop strain, and rms (80dB) vs time; (b) rms (86 dB) vs hoop strain, hoop stress vs hoop strain, and axial stress vs axial strain (rms background = 0.008 V at 80 dB and 0.015 V at 86 dB). For 7075-T651 from reference 3.

Figure 2: Typical test results for a tensile test to approximately 8% total axial strain of the biaxial specimen: (a) axial strain, hoop strain, and rms (100 dB) vs time; (b) rms (86 dB) and axial stress vs axial strain (rmsbackground = 0.019 V at 86 dB and 0.095 V at 100 dB). For 7075-T651 from ref. 3.

Figure 3: Acoustic Emission (rms) above background as a function of strain for a through-the-thickness 7075-T651 tensile specimen. Band pass filter frequency range 100-300 kHz, 86 dB gain.

Figure 4: Plots of AE vs. hoop strain which show comparable high amplitude burst type but dissimilar continuous acoustic emission for air melt specimens, 1 & 2, as opposed to the vacuum melt specimens, 3 & 4, which exhibited burst type and continuous AE of similar amplitude and frequency.

Figure 5: Plots of AE (rms) and stress as a function of strain illustrating the extremely low levels of acoustic emission for air melt, (5) & (6), and vacuum melt, (7) & (8), 21-6-9 when pulled in the longitudinal direction.

Figure 6a: Photograph of the tensile test apparatus for the biaxial test specimen.

Figure 6c: Photograph of internal pressurization test apparatus for the biaxial test specimen.

Figure 6b: Drawing of AE biaxial test specimen (all dimensions are in millimeters, except for the thread size and surface finish which are in inches).

Figure 7: Dimensions of standard tensile specimens for 7075 and 21-6-9.

Table 1: Chemical composition (wt%) of the 7075-T651 aluminum alloy.

Element	This experiment's specimens	Standard specification for the 7075-T651 alloy
Si	0.2	0.5 maximum
Fe	0.22	0.7 maximum
Cu	1.4	1.2 to 2.0
Mn	0.02	0.3 maximum
Cr	0.15	0.18 to 0.40
Zn	5.6	5.1 to 6.1
Ti	0.03	0.20 maximum
Mg	2.3	2.1 to 2.9

Table 2: Chemical composition (wt%) of the 21-6-9 stainless steel alloy.

Element	Air Melt Specimens		Vac. Melt Specimen	Industrial specification
	#1	#2	#4	for the 21-6-9 from Armco
C	0.024	0.029	0.021	0.08 max
Mn	8.2	8.2	9.1	8.00 to 10.00
Si	0.22	0.22	0.45	1.0 max
P	0.03	0.019	0.013	0.060 max
S	0.016	0.016	0.003	0.030 max
Cr	20.0	20.0	20.2	19.00 to 21.5
Ni	6.3	6.3	6.5	5.50 to 7.50
N	0.35	0.35	0.30	0.15 to 0.40
Mo	0.2	0.2	0.06	-
Au	0.011	0.011	0.001	-
Ca	0.01	0.01	0.01	-
Al	0.1	0.1	0.1	-
Ti	0.019	0.019	0.003	-
Fe	bal.	bal.	bal.	balance

Figure 8: Micrographs showing grain structure (a) and inclusions (b) in three perpendicular planes taken from the reduced section of a biaxial specimen of 7075-T651 Aluminum alloy (after a pressurization test). (55x).

Fe

Cu

Mg

Si

Zn

Al

Figure 9: Microprobe analysis of dark inclusion in 7075.
(1 cm = 5 microns)

Fe

Cu

Mg

Si

Zn

Al

Figure 10: Microprobe analysis of light inclusion in 7075.
(1 cm = 5 microns)

Figure 11: A plot of % light inclusion cracking versus ϵ for 7075-T651 specimens subjected to internal pressurization. Percentage cracking corresponds to the number of cracks observed divided by the number of inclusions; can be greater than 100%.

Figure 12: Optical micrographs of inclusions in 7075-T651 where the matrix has been pressurized to 0.8% hoop strain, (a) and (b). The orientation of inclusion cracking with respect to the rolling axis (also specimen axis) and applied load is presented schematically in (c).

Figure 13: (a) and (b). Optical micrographs of inclusions in a 7075-T651 biaxial specimen which had been pulled in tension to 8% strain. (c) Schematic representation of inclusion cracking with respect to rolling axis (also specimen axis) and applied load.

Figure 14: Scanning fractograph of a longitudinal 7075-T651 specimen.
(Tensile and rolling axis are perpendicular to fracture face.)

Figure 15: Scanning fractographs of a longitudinal 7075 tensile specimen showing transgranular ductile tear morphology (a), and an extremely small dimple size in (b). (Rolling direction is perpendicular to surface).

20 x

Figure 16: Fracture surface of 7075-T651 pulled in tension in the through-the-thickness direction. The tensile axis is perpendicular to the surface.

Figure 17: Fracture surface displaying stepped morphology of the 7075 through-the-thickness tensile specimens (SEM's). The step region in (a) is enlarged in (b) and (c) showing the shear wall and plateau areas respectively. Note the cleaved particles in (b) and (c).

a. 26x

b. 145x

c. 1100x

d. 1200x

e. 2100x

f. 2500x

Figure 18: Scanning micrographs demonstrating the complex nature of the fracture morphology of a blaxial 7075-T651 specimen when pressurized to failure. Inclusions are evident in each region. (R.D. = horizontal, hoop direction = approximately perpendicular to each surface.)

Figure 19: Fracture profiles for three combinations of load configuration and specimen orientation examined in this paper: (b) tensile/longitudinal, (a) tensile/through-the-thickness, and (c) pressurized biaxial. (Rolling axis is horizontal.)

Crack initiation and propagation in 7075-T651 aluminum under transverse tensile loading:

1. the matrix in through-the-thickness direction is placed in tension
2. particles crack perpendicular to stress axis at low strain (yield)
3. cracks in adjacent particles of same stringer link by following grain boundary
4. cracks in adjacent stringers join via transgranular link (T)
5. critical crack size is reached -- failure.

Figure 20: A proposed model for the initial stages of critical crack initiation and propagation in 7075-T651 for the given loading conditions. This light micrograph is of a through the-thickness tensile specimen.

Figure 21: Micrographs showing grain structure (a) and inclusions (b) in three perpendicular planes taken from the reduced section of a biaxial specimen of air melt 21-6-9 stainless steel (after a pressurization test). (55x)

Figure 22: Micrographs showing grain structure (a) and inclusions (b) in three perpendicular planes taken from the reduced section of a biaxial specimen of vacuum melt 21-6-9 stainless steel (after a pressurization test). (55x)

INCLUSION CHARACTERIZATION

SPM#	TYPE/SIZE	LENGTH TO WIDTH RATIO/ SEGREGATION TENDENCY	COMPOSITION WT. %	DEGREE (%) OF CRACKING OR DECOHESION	MICRO- HARDNESS*	APPROX. NUMBER (DENSITY IN NO./cm ³) IN MATRIX/ COMMENTS
#1 AIR MELT	A. 3-10 μ m some larger	1:1	-found in stringers and globules	Mn-27Cr-17Ti(Ox)	20-30% cracked thru 3% hoop; mainly lg. inc. No DC (decohesion)	(?) 10 ² (10 ⁴) smaller inclusions 10 ³ (1) lg. globules
	B. 2-5 μ m x 10-20 μ m	4:1	-solitaires, occasionally mixed w/ type A	Mn-50S	No cracking No detectible DC	170kp/mm ² 10 ⁵ (10 ³)
	C. 10-20 μ m x 75-200 μ m	10:1	-idiomorphic x-stals bridged by darker segregation	47Cr-25Mn-7Mg (Nitr.) bridge has Si also	High degree of cracking-300%, parallel to forging dir.	1500-1700 kp/mm ² 10 ² (.1)
#2 AIR MELT	A. ...	...	...	...	...	identical to type A for spm 1 except lg. globules were neither as lg. nor as frequent as in 1
	B. ...	...	...	...	...	same frequency and distribution as in spm 1
#3 VACUUM MELT	D. 1-50 μ m	1:1	-no stringers but larger inc. tended to group smaller inc. randomly dispersed ('clean' mat.)	Mn-35Si-7Mg-30Cr-10Ca(silicate)	small degree of cracking <10%	750 kp/mm ² 10 ³ -10 ⁴ inc. >10 μ m (10)
	B. ...	...	...	...	...	10 ³ (1)
#4 VACUUM MELT	D. ...	...	...	...	...	same as in 3
	B. ...	...	...	...	...	10 ³ (1)

*From ref. 21

Figure 23: Chart of inclusion types and their respective characteristics observed in 21-6-9 air and vacuum melt specimens.

Figure 25: Light micrographs of the inclusion types as defined in Table 3 found in air melt 21-6-9 stainless steel. Shown in (a) are examples of type A (Ti) in stringer form and a sulphide particle (S), type B, with common oblong shape. Type C is depicted in (b) where the Ti-Cr-Mn oxide (Ti) globules are bridged by Mn-Si complex (Si). Note high degree of cracking in (b). (From reduced section of biaxial specimen after pressurization).

Figure 26: Light micrographs of inclusions in air melt 21-6-9. Typical stringer grouping of Type A is illustrated in (a) while in (b) cracks parallel to the inclusion axis (bar axis) are evident as are scavenging sulfide particles (S). (From reduced section of biaxial specimen after pressurization).

Figure 27: Light micrographs of inclusion type C in air me + 21-6-9. This inclusion type is characterized by large + titanium rich globules (Ti) separated by Mn-Si oxide bridges (S) in (b). (From reduced section of biaxial specimen after pressurization. Bar axis is parallel to inclusion axis).

Figure 28: Optical and scanning micrographs of air melt 21-6-9 showing distribution of important elements in inclusions (a) and crack orientation (b).

a.

100 μm

b.

20 μm

Figure 29: Optical micrographs illustrating the morphology (color, shape, & size) and nature of segregation of Type D inclusions found in vacuum melt 21-6-9. (From reduced section of biaxial specimens after pressurization. Bar axis is parallel to inclusion axis).

Figure 30: Light micrograph of a nitride particle, type D, from air melt 21-6-9 illustrating substantial cracking; a result of tensile loading.