ACOUSTIC EMISSION STUDIES OF ELECTRON BEAM MELTING AND RAPID SOLIDIFICATION

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ABSTRACT

Acoustic emission is well suited for monitoring electron beam induced melting and rapid solidification, providing a real-time, volumetric survey of the process. The present study examines the origin of acoustic emission during melting and rapid resolidification of Al and Al-4.5% Cu using heat flow theory and microstructure characterization. Specimens were heated to a steady state using a pulsed electron beam and acoustic emission measured during heating and subsequent cooling. The heat flux level was systematically increased so that both solid state heating and melting could be independently studied.

The electron beam, because of its relatively slow rise time, was found not to produce acoustic emission directly. The motion of the liquid-solid interface, for the same reason, also fails to generate detectable acoustic emission. Acoustic emissions were found to be generated by plastic deformation and sometimes crack growth. The emission was strongly influenced by the heat flux level (ultimate temperature), prior cold work, alloying, and heat treatment. A thermal stress model was used to develop an understanding of plastic deformation/fracture initiation and propagation during the heating and cooling cycle.

INTRODUCTION

Directed high energy sources (lasers and electron beams) are currently being evaluated for rapid surface melting and resolidification of a range of engineering alloys (1). In these processes, the bulk substrate, in intimate contact with an electron beam melted surface layer acts as the quenching medium. This results in high liquid-solid interface velocities and short solidification times during cooling after electron beam heating has terminated (2,3). These rapidly solidified layers exhibit enhanced wear and corrosion resistance arising, it is believed, from improved chemical homogeneity and the fine scale of the microstructure (1). These materials offer promise for applications such as bearing surfaces, cutting tool faces and corrosion-resistant parts.

At present, critical process variables such as liquid-solid interface velocity or the cooling rate can only be deduced from the one- and two-dimensional heat flow models of melting and resolidification. There are no in situ methods of monitoring the important process variables. If measured, they could provide feedback/feedforward information for in-process control of microstructure and the elimination of flaws: a real-time monitoring and diagnostic technique is needed.

Acoustic emission (AE) is an attractive potential candidate for such an application. However, the origin of acoustic emission during electron beam melting solidification is not well established. When AE is correlated with heat flow theory and microstructural examination, considerable light can be shed on operative defect mechanisms, such as plastic deformation or crack initiation and propagation. This paper illustrates the application of this method to pure Al and Al-4.5 wt.% Cu.

Heat Flow Theory. Heat flow for finite diameter electron beams can be modeled using two-dimensional heat flow theory. Results of calculations for two-dimensional heat flow for rapid surface solidification in pure Al are given in Ref. (3). Similar calculations for Al-4.5% Cu are given in Ref. (8,9). The results indicate a two-dimensional steady-state heat flow is ultimately attained when the heat absorbed over the circular region exposed to the beam is exactly offset by heat conduction into the substrate interior. At steady state, the maximum temperature is located at the surface at the center of the beam spot (T(0,0)); the thermal profile then remains fixed (with only static stresses) from which solidification proceeds.

Two dimensionless variables are used
FIG. 1. Temperature at the center of the circular region during melting of a semi-infinite Al-4.5 wt % Cu alloy substrate as a function of uniform absorbed heat flux ($q_a$), radius of the circular region ($a$) and time ($t$) (8).

Table 1
Absorbed Uniform Heat Flux to Reach Critical Temperatures (2,3,8,9)

<table>
<thead>
<tr>
<th>Phenomenon</th>
<th>Temperature °C</th>
<th>Absorbed heat flux*, $q_a$ (W/m)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Al</td>
<td>A1-4.5% Cu</td>
</tr>
<tr>
<td>Melting initiated</td>
<td>660</td>
<td>536</td>
</tr>
<tr>
<td>Melting completed</td>
<td>660</td>
<td>647</td>
</tr>
<tr>
<td>Vaporization</td>
<td>2447</td>
<td>2368</td>
</tr>
</tbody>
</table>

*Assumes initial temperature to be 24°C.
to describe heat flow in the solid state:

\[
\text{Dimensionless temperature} = \frac{k_s(T(0,0)-T_0)}{qa}
\]

\[
\text{Dimensionless spot size} = \frac{a}{4\sqrt{\alpha t}}
\]

where \(k_s\) is thermal conductivity of the solid, \(T_0\) is the initial substrate temperature, \(q\) and \(a\) are the absorbed heat flux and electron beam radius, \(\alpha\) is the thermal diffusion coefficient of the solid, and \(t\) is the time measured from the instant the beam was switched on. When the dimensionless temperature equals 1, steady state is established. If \(T(0,0)\) is equal to the melting point or solidus temperature (for an alloy), the product \(qa\) corresponds to that above which melting will initiate at the beam center. As \(qa\) is increased, melting will spread outward from the center of the spot to cover the entire circular area. This is reached when

\[
(k_s(T(0,0)-T_s) + \Delta \alpha + k_s(T_s-T_0))/qa = 1
\]

where subscripts \(i\) and \(s\) stand for liquid and solid, respectively, and \(T_i\) and \(T_s\) are the liquidus and solidus temperatures. \(\Delta \alpha\) is a term accounting for heat conduction in the "mushy" region and is calculated to be \(6.2\times10^4\) W/m in Al-4.5% Cu (8).

Figure 1 shows the relationship between dimensionless temperature, beam radius and time during heating in the alloy (8). (The corresponding curve for pure Al is given in Ref. 3). Using these figures one can calculate the \(qa\) level at which melting initiates, that at which it fills the beam spot and that for vaporization to occur. (Table 1).

In designing the acoustic emission experiment, our first concern has been to allow sufficient time during heating for steady state temperature to be obtained so that a constant stress is reached before solidification commences. For a given beam radius \(a\), the time to reach steady state can be taken as that at which \(\frac{a}{4\sqrt{\alpha t}}<0.5\), as can be seen in Fig. 1, or \(t_s>\frac{a^2}{\alpha}\). For both materials this is approximately 13 ms. Our use of a 77 ms pulse meets the requirement for attainment of steady state conditions.

Our second concern is that, during cooling, the measurement period should extend over the cooling time of the melted region. Using two-dimensional heat flow equations (3), we examined the thermal effects of a 77 ms pulse in pure Al heated to 650°C. Following the heat pulse, it cooled to 114°C within 20 ms, and to 45°C within 100 ms. We used a 77 ms measurement period which covers the majority of the cooling period.

Microstructure. Microstructure, temperature and thermal stress determine the defect mechanisms responsible for the acoustic emission. Microstructural investigation is thus crucial to an understanding of the origin of the acoustic emission.

An example from previous work (10) can illustrate this point. We consider the effect of increasing the product \(qa\) upon melt depth and microstructure for commercially pure 1000 aluminum and an Al-6.5% Cu alloy (Figs. 2-4). The melt depth increased with increasing \(qa\); the alloy, because of its lower melting temperatures, had a greater depth of melt than the 1100 aluminum. A comparison between the calculated and experimentally measured melt depths was made (10). Agreement was good considering various assumptions of the heat flow model including constant thermophysical properties. The alloy had a less clear transition from unmelted to melted regions, exhibiting a " mushy" zone of partially melted material (Fig. 3). It was possible to tentatively link the source of emission to crack production in the alloy, since much greater signals were obtained in the alloy, which cracked, than the pure Al, which did not crack. There were, however, shortcomings with these initial experiments which are addressed below.

Acoustic emission. In this phenomenon, stress waves are emitted during rapid defect motion or phase transformations. The elastic waves then propagate throughout the structure to be detected by remotely located transducers. For a recent review of the potential for process monitoring see Ref. 4. The technique has a real-time capability (limited only by the time for elastic waves to travel from the source to receiver, usually a few microseconds) and has been successfully applied to monitoring commercial welding (5,6,7). In the present work, we report the use of acoustic emission to monitor rapid solidification of electron beam surface melted pure Al and Al-4.5 wt.% Cu. Heat flow theory has been used to design controlled experiments in which shallow melts were formed at the surface of aluminum plates. After attainment of steady-state temperatures the electron beam was switched off, resulting in rapid solidification. Acoustic emission was measured separately during both heating and cooling since the origin of the emission was different in these stages. It is important to understand both since they jointly contribute to the instantaneous emission during (continuous) surface modification treatments or welding. The effects of alloying, cold work, and heat treatment were systematically studied as a function of heat flux level (i.e. ultimate steady-state temperature). Post mortem metallography and heat flow theory were correlated with acoustic emission to make tentative identifications of the defect mechanisms accompanying heating and cooling.

The present study was undertaken to gain a better understanding of the relative contributions to the detected acoustic emission of dislocations and cracks and how this is affected by microstructure. High purity Al and a binary Al-4.5% Cu alloy were used, rather than the commercial materials used in a previous study because these are less complex and maybe better charac-
FIG. 2. Optical micrographs of 1100 Al and 2219 Al alloy (6.5% Cu) showing the effect of qa on the melt profile and ensuing microstructures in the "steady-state" region.

FIG. 3. Optical micrographs of 2219 Al alloy (6.5% Cu) melted at qa = 2.7×10^5 W/m showing detailed microstructure of the "mushy" zone at the bottom of the melt pool.
terized. In addition, previously published heat flow calculations (8) could be more rigorously applied. Also previous work (10) over a wider qa range (temperature range) encountered problems associated with vaporization. This complicated the analysis and has been excluded here. The measurement system has been improved reducing signal distortion and improving the signal-to-noise ratio, and a new specimen/transducer holder design has improved the measurement reproducibility.

EXPERIMENTAL PROCEDURE

Materials. In this study, effects of substrate composition and condition were examined by using pure Al in an annealed and 20% cold-worked state and an Al-4.5 wt.% Cu alloy in two heat treated conditions.

Both materials were at least 99.99% pure and after casting were homogenized at 535°C for 1.5 hours, hot rolled from that temperature, solution treated at 535°C for 75 min. and ice water quenched. The alloy composition was verified to be 4.5±0.1 wt.% Cu (at four locations) by wet chemical analysis. The specimens were then machined to 0.4x2.5x2.5 cm squares from the annealed rolled plate. Half of the pure Al specimens were further cold rolled 20% in thickness. The alloy specimens were then given low temperature heat treatments: Group 1) aged to peak hardness (8 hours at 179°C) and Group 2) underaged (1 hour at 170°C). The final grain size was approximately 450 microns in the alloy and 350 microns in the pure Al (at the surface, whose 1 mm depths had recrystallized). The specimens were stored in liquid nitrogen to avoid the possibility of precipitation between stages of the above heat treatment and testing.

Electron Beam Operation. Figure 5 is a schematic diagram of the electron beam facility, together with the acoustic emission apparatus and associated instrumentation. The sample and holder were placed in the electron beam apparatus and evacuated to 10⁻⁵ Torr. The beam was accelerated across 22 kV and focused to 1 mm radius spot (with an approximately rectangular intensity profile). The absorbed heat flux was determined at each flux level by measuring the current flowing through a resistor between the specimen and ground. The error in qa was estimated to be approximately ± 15%. A series of single 77 ms duration pulses spots of increasing qa level were applied to the surface of the samples. The beam and sample were stationary so that acoustic emission during heating was clearly separable from that of cooling.

Acoustic Emission Measurements. The acoustic emission measurement system is shown schematically in Fig. 5. A piezoelectric, 140 kHz resonant acoustic emission transducer was coupled to the back of the specimen with a low viscosity oil. To minimize ringing, which prolongs the period of emission and may cause obliteration of closely following signals, the specimen was covered on the underside with a plastic adhesive tape and held down by an O-ring. Spring loading insured constant coupling for improved measurement reproducibility. Bulk specimen heating was less than 5°C degrees per pulse, and since generally no more than eight pulses were applied to a sample, its bulk temperature rise was <40°C. An acoustic emission parameter analogous to energy was measured by summing the squared unbiased voltages of digitally recorded signals for periods of 77 ms during both heating and cooling.

RESULTS

A typical electron beam pulse and superimposed AE signal are shown in Fig. 6. Different AE signals had several common characteristics. There was a delay time after the beam was switched on before detectable emission occurred. This delay decreased with increasing qa (i.e., increasing temperature). Acoustic emission was then produced during heating. After the the beam was switched off, there was a delay time prior to emission provided that T(0,0), the temperature at the center of the heated region, approached or exceeded melting (12). Then "cooling" acoustic emission occurred.

Acoustic emission on heating was relatively small for T(0,0) < Tₘ, the solidus temperature. Above Tₘ, it increased dramatically (Figs. 7 and 8) for both materials. Acoustic emission on cooling was smaller than that produced during heating at temperatures below Tₘ. The "cooling" acoustic emission had three characteristic temperature (qa) regions indicated in Table 2 and Figures 7 and 8. In region I, the emission value increased with temperature (qa) up to a maximum value. It then decreased in region II to a vanishing value at the qa level at which melting was initiated. It then increased again in region III, during general melting.

Typical microstructures of the heated zones are shown in Figures 8-10. At the lower temperatures, corresponding to region I, all of the specimen surfaces showed coarse slip bands (Figs. 8A and 9A). The amount of deformation (number and width of slip bands) increased with increasing temperature (qa).

Although some recovery, as evidenced by subgrain boundary development, is seen at the lowest heat flux levels in pure Al (Fig. 8A), it predominated in region II. Region II corresponds to temperatures in which recovery processes occurred, a typical example of which is shown in Fig. 9A for the Al-4.5% Cu alloy.

No cracks were observed at any level of qa (temperature) in the pure Al samples (although the grain boundaries were thermally etched, especially at the higher temperatures, giving the appearance of cracks, Fig. 8B). Fig. 8C shows a typical cross-section of a melt zone in pure Al. There are two grains visible, but no fractures.

In region III, where general melting
FIG. 4. The effect of $q_a$ upon melt depth for 1100 and 2219 aluminum alloy. Melt depth is measured from the liquidus isotherm to the substrate surface.

FIG. 5. Schematic diagram of electron beam apparatus and acoustic emission instrumentation.
FIG. 6. Typical acoustic emission signal produced by melting and resolidification of surface melt in peak-aged Al-4.5%Cu alloy. (A) During heating at $q_a = 8 \times 10^4$ W/m and (B) During cooling after beam is switched off.
FIG. 7A. Acoustic emission energy generated during surface heating and cooling as a function of $q_a$ level in pure Al. Data points an average of 8. Standard deviation is $\pm$ 43% on heating and $\pm$ 22% on cooling.

FIG. 7B. Acoustic emission energy generated during surface heating and cooling at given $q_a$ level in Al-4.5% Cu alloy. Data points an average of 8. Standard deviation is $\pm$ 43% on heating and $\pm$ 43% on cooling.
FIG. 8A. Annealed pure Al. Optical micrograph on surface at center of irradiated area ($q_a=4 \times 10^4$ W/m$^2$). Surface polished prior to radiation; not etched. Notice recovery: sub-boundary development at center surrounded by coarse slip bands confined to grains.

FIG. 8B. Annealed pure Al. Optical micrograph of surface melt ($q_a=1.4 \times 10^5$ W/m$^2$); surface polished prior to irradiation; unetched. Gray crescent to right of melt zone is surface upheaval caused by melting. Note radial thermal grain boundary grooves in resolidified melt, frozen-in surface waves, recovery sub-boundaries, coarse slip.

FIG. 8C. Annealed pure Al. Optical micrograph section through surface melt ($q_a=1.4 \times 10^5$ W/m$^2$) similar to that above. Note that there are no cracks. Electropolished and Keller's Etch.
FIG. 9A. Underaged Al-4.5% Cu. Optical micrograph of surface (polished prior to irradiation) at center of irradiated area (qa=6x10^4 W/m) showing coarse slip bands confined to grains.

FIG. 9B. Underaged Al-4.5% Cu. Optical micrograph of surface (polished prior to irradiation) (qa=8x10^4 W/m) showing partial melt at center of irradiated area, surrounded by circle of recovery, in turn surrounded coarse slip bands.

FIG. 9C. Underaged Al-4.5% Cu. Optical micrograph of section through resolidified melt (qa=1.6x10^5 W/m) showing "hot tearing" cracks there. Keller's etch.
Fig. 10A. Peak-aged Al-4.5% Cu.
Optical micrograph of section through center of irradiated region (qa=8×10⁴ W/m) showing sub-boundary development. Keller's etch.

Fig. 10B. Peak-aged Al-4.5% Cu.
Optical micrograph through section of regolidified melt (qa=2×10⁵ W/m). Keller's Etch.

Fig. 10C. Peak-aged Al-4.5% Cu.
Expanded optical micrograph of the above showing resolidified intergranular "liquation" in substrate. Keller's etch.
occurred, the difference in microstructures of the resolidified melt zones of pure Al and the 11-4.5% Cu alloy is dramatic. In the resolidified melt zone of the alloy material, numerous "hot tear" fractures are observed in the melt zone cross section (Figs. 9c, 10b). These appear, at this magnification, to be intergranular cracks extending into the substrate. In fact, at higher magnification (Fig. 10c) it is clear they are not cracks but melting and resolidification of grain boundaries. Savage (16) first described this phenomenon, calling it "constitutional liquation", and the term "liquation" has since come into general usage. It occurs in eutectic alloys and is associated with microsegregation at the grain boundaries that preferentially melt when heated above the eutectic temperature (16). Intergranular liquation occurred for all of the alloy melts along all of the grain boundaries interfacing the melt zone.

A summary of these observations is given in Table 2.

**DISCUSSION**

Thermal Stress Cycle. When the circular region of the beam spot is heated and then cooled, material is subjected to a thermal stress cycle. Phase transitions, plastic flow and cracking can further complicate this stress state. To determine the origin of the acoustic signal it is essential to have some understanding of the sequence of stresses. A semi-quantitative understanding can be obtained using the concept of "stress-free" strain developed by Eshelby for study of the "inclusion problem" (13).

Figure 11 schematically depicts the thermal cycle. The horizontal axis represents temperature in the disc (the plate always remains at room temperature) and the vertical axis time. When the temperature profile is invariant then it is possible to approximate the heated region as a disc of radius a in a plate. i.e., the temperature profile is approximated as a top hat distribution of width 2a. In step 3 the disc is imagined to be cut out of the plate and heated, causing a "stress-free" train. In step 2, forces are applied to the disc so that it once again fits into the original hole. The interface is welded, and the forces removed, causing "thermal" stresses (depicted by arrows) in the disc and adjacent matrix. It can be shown that the radial stresses are compressive everywhere. Tangential stresses are compressive within the disc and tensile without (13). The compressive radial stresses inside the disc may exceed the yield strength of the hot material, resulting in plastic deformation (coarse slip) and rumpling of the surface. Since there will be no tendency for cracking inside the disc during heating (because of compressive stresses), cracks (or intergranular separation aided by liquation) can form outside the disc.

**FIG.11.** Two-dimensional representation of thermal stresses generated by pulsed electron beam irradiation of circular region of radius a.

Steps 3 and 4 correspond to stress relaxation in the steady-state condition by a number of mechanisms including melting, recovery (subgrain boundary formation), creep, crack growth and intergranular liquation. In the limit the disc and plate ultimately become once more stress-free.

In step 5 (cooling) we imagine the now stress-free but hot disc of radius a to again be cut from the plate. The disc is cooled, resulting in a stress-free strain, and then reinserted into the plate as before. The resulting thermally-induced stresses produced by cooling are opposite to those produced by heating. Accordingly there are tensile stresses in the radial directions everywhere, and the tangential stresses are tensile inside the disc and compressive outside. This biaxial tensile stress state produced inside the disc during cooling is conducive to crack formation there during and following solidification. Only one tensile stress (in the radial direction) is induced outside the disc during cooling. This could promote tangential cracks (i.e. cracks containing the tangential direction) during cooling, but these are not seen. These cracking processes and plastic deformation partially relax the stress during cooling.

Acoustic Emission Mechanisms. The heat flow theory discussed above indicates that, for electron beam heating, the rise and decay...
time of the temperature is on the order of 10 and 100 ms, respectively. As a result, the rise time of the thermal expansion stress corresponds to a frequency of <100 Hz, far too low to be detected as acoustic emission. Hence the electron beam heating itself cannot be the source of the emission we observe. (This is not generally true of Q-switched laser pulses, which are well-known generators AE signals (17)). This is experimentally supported by the existence of a delay time for acoustic emission on heating. Heat flow theory likewise indicates motion of the liquid-solid interface occurs typically over 10-100 ms, so that it too can be eliminated as a direct AE source. Other processes, such as radiation pressure and ablation (vaporization) are calculated to be far below the detection sensitivity of the acoustic emission system (12).

Microstructural examination, as summarized in Table 2, suggests that plastic flow and/or fracture may be the principal sources of AE. All of the material underwent some amount of coarse slip, which is known to produce significant emission. Madsen et al. (4,14) have shown that acoustic emission can be detected provided that the product na v > 10^-3 m^2 s^-1, where n is the number of moving dislocation segments, a is the slip distance, and v is the velocity. Consider the early stages of slip where dislocation loops can expand across the entire grain before being arrested at the boundaries. If we take n=1, a=200 μm (the grain radius), and estimate v=200 ms^-1, we obtain na v=0.04, which is over 10 times the detection threshold. Thus a single high speed dislocation loop could give a detectable signal.

During heating, as the steady-state temperature increases, the microstructure remains relatively fixed, so that n and a will be relatively constant. However, increasing thermal stress (increasing qa) and enhanced thermal activation are expected to increase the dislocation velocity. The volume of stressed material also increases with temperature (qa). Thus, we can expect the acoustic emission to increase with qa in region I in accordance with observations. For heating, the same processes continue in regions II and III. In addition, bulk and intergranular melting, with associated stress relief, also occur. This could account for the increasing AE with increasing qa on heating and its dramatic increase above the solidus temperature (Figs. 7A and 7B).

The picture is different during cooling because of resolidification and the annealing (recovery) of deformed material in the heat affected zone. One tentative explanation is as follows. In region I, where the temperature is lowest, these effects will be minimized, and deformation processes will be somewhat reversible, so that the AE on cooling increases in region I through the increased stress and volume of material cooling. In region II, recovery processes will be enhanced with increasing qa. Recovery results in a relaxation of stress and development of subgrain boundaries. The subgrains limit the distance dislocations move so that AE on cooling in region II decreases with increasing qa. In region III, however, the amount of material affected by the melting and resolidification process (which increases with qa) predominates over other effects to bring

<table>
<thead>
<tr>
<th>Region</th>
<th>qa (W/m)</th>
<th>T(0,0) (°C)</th>
<th>Final Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al I</td>
<td>&lt;0.6 x 10^5</td>
<td>&lt;250</td>
<td>Coarse slip, sub-boundary development, no melting.</td>
</tr>
<tr>
<td></td>
<td>0.6-1.4 x 10^5</td>
<td>250-660</td>
<td>Heavier coarse slip, sub-boundary development, no melting</td>
</tr>
<tr>
<td></td>
<td>&gt;1.4 x 10^5</td>
<td>&gt;660</td>
<td>Melting, coarse slip, sub-boundary development.</td>
</tr>
<tr>
<td>Al-4.5%Cu I</td>
<td>&lt;0.6 x 10^5</td>
<td>26-250</td>
<td>Coarse slip, no melting</td>
</tr>
<tr>
<td></td>
<td>0.6-1.0 x 10^5</td>
<td>250-560</td>
<td>Coarse slip, sub-boundary development no melting.</td>
</tr>
<tr>
<td></td>
<td>&gt;1.0 x 10^5</td>
<td>&gt;560</td>
<td>Melting, coarse slip, sub-boundary development, intergranular liquation, hot tears in resolidified melt.</td>
</tr>
</tbody>
</table>
about an increase in acoustic emission.

The condition of the substrate will modify these processes. Cold work limits the distance dislocations may move during subsequent plastic flow. This, according to our model, is expected to reduce the AE intensity as observed in these experiments. In Al-4.5% Cu undergoing produces only partially developed GP zones (15) which promote coarser slip compared with that of a solid solution or peak-aged structure (14). This results in an increase in acoustic emission (over that of a solid solution) in tensile tests at room temperature. The present tests, however, were (in effect) at elevated temperatures, during which the metastable GP zones are expected to begin to dissolve. Thus coarse slip may be suppressed and with it intense AE.

CONCLUSIONS

Table 2 summarizes many of the results responsible for AE at various levels of heat flux (Regions I, II and III). We can draw some general conclusions about the mechanisms involved in acoustic emission during rapid surface melting and resolidification.

1. Processes that are directly associated with electron beam heating generally vary relatively slowly with time. Their elastic wave radiation is thus at frequencies too low to directly cause observable acoustic emission. Thus, electron beam heating, radiation pressure, substrate quenching, and motion of the liquid-solid interface itself are not direct causes of detected acoustic emission.

2. However, relaxation of these stresses by defect motion (i.e., slip or cracking) can occur with sufficient speed to generate high frequency elastic waves and is postulated to be the origin of AE.

3. The condition of the substrate (its composition, heat treatment or degree of cold work) strongly affects the amount of emission by controlling the micromechanisms of slip.

4. The amount of acoustic emission is greater on heating than on cooling, at least for the materials studied here, increasing exponentially above the solidus.

If generally true, it will be necessary to spatially filter data during AE monitoring of welding or surface modification in which the heat source is moving, so that the emission which occurs on cooling may be unmasked from that of the region being heated.

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