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Magnetoacoustic and Barkhausen Emission in Ferromagnetic Materials [and Discussion]

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Magnetoacoustic and Barkhausen emission in ferromagnetic materials

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Magnetoacoustic emission (MAE) and Barkhausen emission (BE) have been studied in ferromagnetic materials placed in a magnetic field, varying at a few millihertz. Comparison of the two signals indicates the nature of the domain walls responsible for the activity at a given field strength. In order to characterize a specimen the strength of the emission around the hysteresis loop is measured.

The technique has been used to measure non-destructively the effects of the following.

(*a*) Precipitates: both MAE and BE are sensitive to the growth of precipitates in Incoloy 904, and can be used to monitor the precipitate size.

(*b*) Dislocations: BE and MAE exhibit high sensitivity to plastic deformation, and this has been studied in α-Fe.

(*c*) Radiation: neutron irradiation of an Fe-Cu alloy produces small changes in emission, although with sufficient sensitivity for useful characterization of radiation effects.

(*d*) Tensile stress: MAE is sensitive to stress over the whole range, but particularly at low stresses. Such measurements lay the foundation for the use of MAE for residual-stress measurements.

1. Introduction

The discontinuous changes in magnetic moment which occur when a ferromagnetic material is placed in a varying magnetic field, can be detected as sharp transient pulses of EMF across a search coil in close proximity to the material. This is known as the Barkhausen effect (Barkhausen 1919), and arises because the magnetic domain walls jump from one pinning point to the next. At high fields this Barkhausen emission (BE) may be due to domain wall creation or annihilation, or to the rotation of domain magnetization. Abrupt domain wall movement may also generate transient elastic waves because of the changes in magnetostrictive strain produced in the volume of material swept out by the domain wall. These elastic waves can be detected by a piezoelectric transducer bonded to the specimen (Lord 1975) and are known as magnetoacoustic emission (MAE).

Two important differences between the sensitivities of BE and MAE to domain wall motion should be noted. First, whereas the BE signal is proportional to $\sin^2 \frac{1}{2}\theta$, where θ is the angle between the magnetization vectors in the two domains on either side of the wall, MAE is proportional to $\sin^2\theta$. Thus BE is a maximum for $\theta = 180$ °C, but MAE is a maximum for $\theta = 90$ ° and zero for $\theta = 180$ °, because 180 ° walls do not give rise to changes in local strain. Secondly, while MAE signals can generally be detected from the bulk of the material, the BE signals are increasingly attenuated at higher frequencies by eddy-current shielding, so that they can only be used to probe near-surface (10–100 μm) effects (Sundström & Törrönen 1979).

BE has been used as a non-destructive examination (NDE) technique to investigate microstructure and stress effects in the laboratory (Tiitto 1977; Donaldson & Pasley 1967; Gardner *et al.* 1971, Altpeter *et al.* 1981). MAE has also been proposed as an NDE technique for microstructural and stress measurement (Kusanagi *et al.* 1979*a*, *b*; Shibata & Ono 1981). It exhibits a greater sensitivity to stress than BE and, provided that the apparatus is sufficiently sensitive to detect the MAE at low magnetizing frequencies, measurements are not restricted to the surface region of the specimen. There is at present a strong motivation to develop NDE techniques such as MAE and BE for residual-stress measurement. However, the relations of both MAE and BE to stress and microstructure are complex and not yet sufficiently well understood to separate stress and microstructure effects. The purpose of the work presented here is to improve our understanding of MAE and BE by systematic studies of the effects of the following.

(*a*) Precipitates: in Incoloy 904, following ageing to grow precipitates (i.e. non-magnetic inclusions) of varying size.

(*b*) Dislocations: in α-Fe, following plastic deformation and subsequent annealing heat treatments.

(*c*) Neutron irradiation: in an Fe–0.2% Cu alloy, where irradiation is known to cause hardening and embrittlement.

(*d*) Tensile stress: in iron, nickel, mild steel and Incoloy 904.

2. Experimental technique

2.1. *Barkhausen-emission activity profiles*

A schematic diagram for the BE measurements is shown in figure 1. A function generator is set to give a 5 mHz triangular waveform output, and this signal is amplified by a 1.2 kW power amplifier which drives the water-cooled magnetizing solenoid. The specimen is in a uniform magnetic field in the centre of the solenoid, and wound around the specimen is a 3450 ± 50 turn detection coil. A similar coil, wound in the opposite sense, is on the same axis

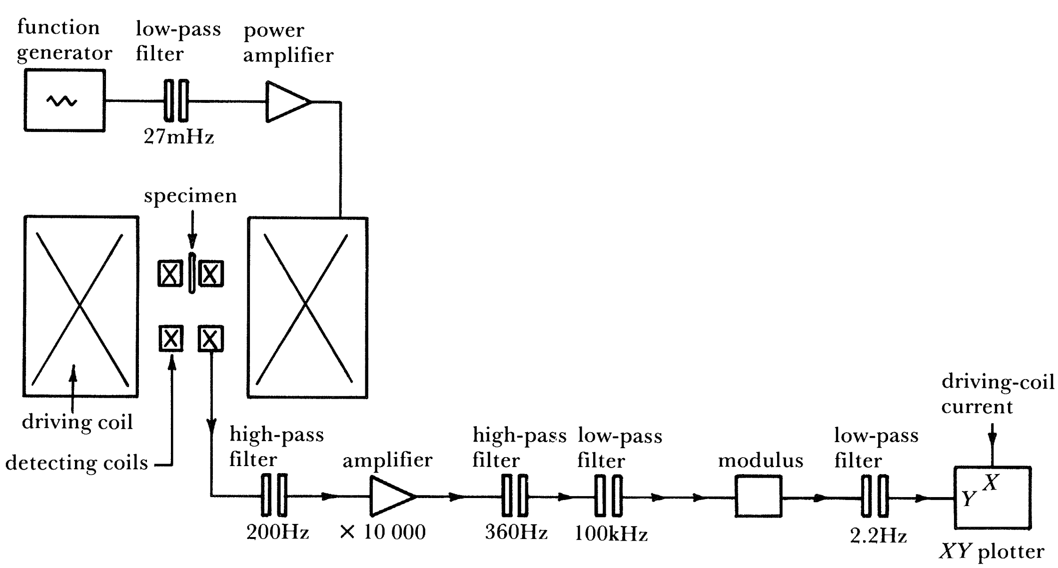

FIGURE 1. Experimental arrangement used to measure BE activity as a function of magnetizing current.

below the specimen and the two coils are electrically connected in series. The EMF induced across these coils (the Barkhausen signal) is suitably filtered and amplified, and the envelope of the signal is plotted on an XY plotter as a function of the current through the magnet. The current range chosen for this experiment was ± 2.4 A corresponding to $H = \pm 21.6$ kA m^{-1}. Figure 2*a* shows a typical result for a single-crystal Incoloy 904 specimen. Note that there are three distinct peaks in the activity, one at low fields and two at the fields corresponding to the knees of the hysteresis loop.

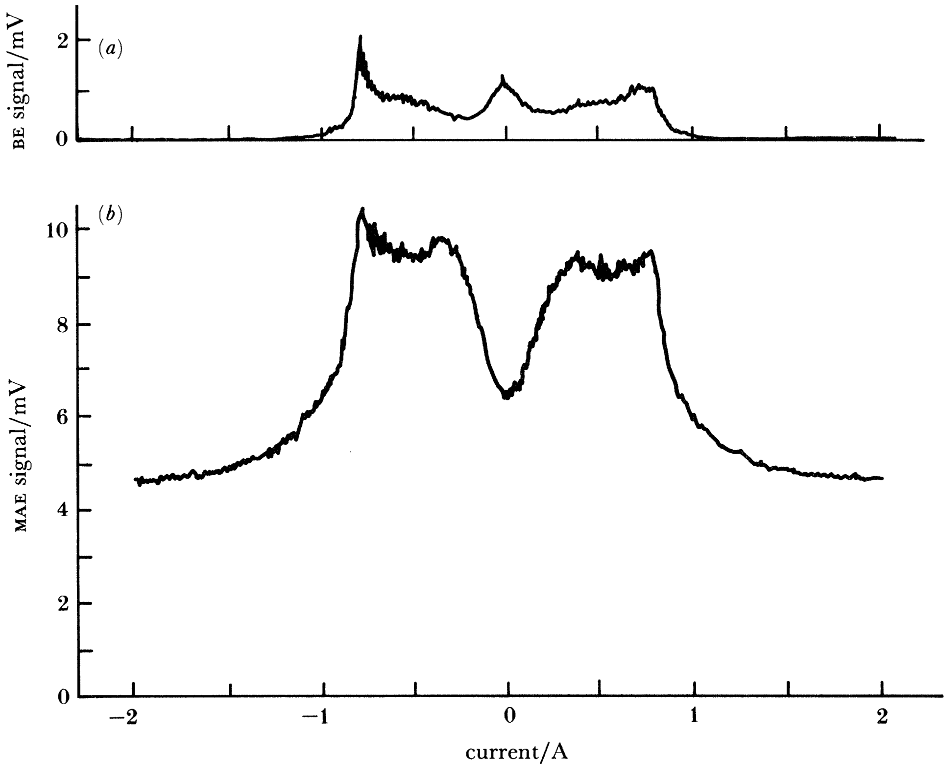

FIGURE 2. Example of (*a*) BE and (*b*) MAE activity profiles from single-crystal Incoloy 904 aged for 355 h at 700 °C.

In the case of the stress measurements, a solenoid was mounted on a 5 kN Instron testing machine and a single detection coil with 3200 turns was used, this coil having slightly higher sensitivity to small flux changes than those mentioned previously. In order to reduce the interference due to the noise generated by the power amplifier in the drive circuit (because this is not rejected as in the two coil system) a 2 A, 45 Hz low-pass filter was placed after the power amplifier. The current range chosen for the stress measurements was ± 1.6 A, corresponding to $H = \pm 24.6$ kA m^{-1}.

2.2. *Magnetoacoustic emission activity profiles*

A schematic diagram for the MAE measurements is shown in figure 3. The output of the sweep device varies linearly from -1.86 V to $+1.86$ V at a rate of 11 mVs^{-1} (1.48 mHz). This signal was amplified to drive the solenoid over the current range of ± 2.4 A. In order to improve the

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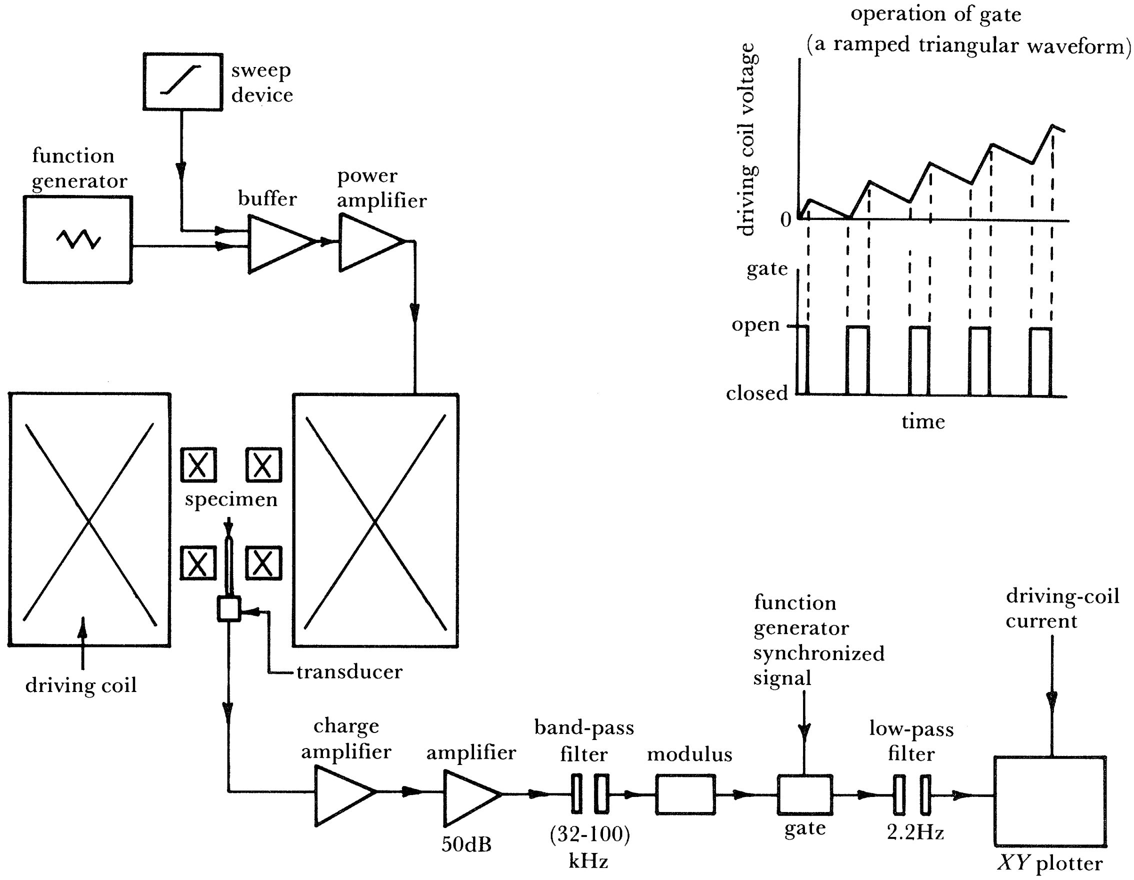

FIGURE 3. Experimental arrangement used to measure MAE activity as a function of magnetizing current.

MAE signal: noise ratio a small 20 Hz (25 Hz in the case of the stress measurement), 150 mA RMS triangular wave was added to the driving signal by using a function generator. The acoustic signal was detected by using a 2.5 MHz damped piezoelectric transducer (sensitivity *ca.* 0.1 $\mathrm{Cm^{-1}}$), and then amplified by using a charge amplifier (with gain of 0.13 V $\mathrm{pC^{-1}}$). The transducer was mechanically attached to the end of the specimen, with Ultragel acoustic coupling fluid. The MAE signal was further amplified by 50 dB and filtered with a bandpass of 32–100 kHz. The signal was then gated at 20 Hz (or 25 Hz) in order to remove the contribution due to the reverse part of the modulation field, and the signal envelope plotted on an *XY* plotter as a function of the magnet current.

In order to check the acoustic coupling a 1.00 A RMS 20 Hz driving signal was used and the voltage at the output of the full-wave rectifier was monitored. The transducer and specimen were then adjusted slightly to obtain the maximum acoustic signal.

Figure 2*b* shows the MAE activity profile for the same single-crystal Incoloy 904 specimen as before. Note that there are only two peaks in the profile, which occur at the same fields as the two outer peaks in the BE activity profile. This implies that the central peak in the BE activity profile mainly arises from 180° domain-wall movements.

3. Investigation of precipitation in Incoloy 904

Incoloy 904 (51.0% Fe, 33.8% Ni, 14.0% Co, 1.2% Ti by mass) is a high-strength, low-thermal-expansion alloy used in machine components. The purpose of this experiment is to show how BE and MAE can be used to monitor the growth of precipitates. By isothermally ageing the alloy, a random distribution of spherical precipitates (i.e. non-magnetic inclusions) can be induced to grow. The volume fraction of these precipitates remains constant but their diameter can be varied by the length of the heat treatment.

Twenty cylindrical polycrystalline specimens were solution treated at 1050 °C in vacuum for 2 h and water quenched. They were then aged at 700 °C in vacuum in order to grow the non-magnetic inclusions, the ageing times ranging from 0–1730 h. The times were chosen on the basis of the Lifshitz–Wagner theory of precipitation, which proposes that the diameter of the inclusions is proportional to the cube root of the ageing time ($t_a^{\frac{1}{3}}$) (Lifshitz & Slyozov 1961; Wagner 1961) i.e. 1, 8, 27, 64, ... h.

The BE activity profiles for these specimens have three peaks similar to the single crystal specimen profile shown in figure 2*a*. A plot of the peak heights as a function of the parameter $t_a^{\frac{1}{3}}$ is shown in figure 4*a*. The height of the central peak rises to a maximum after ca. 300 h

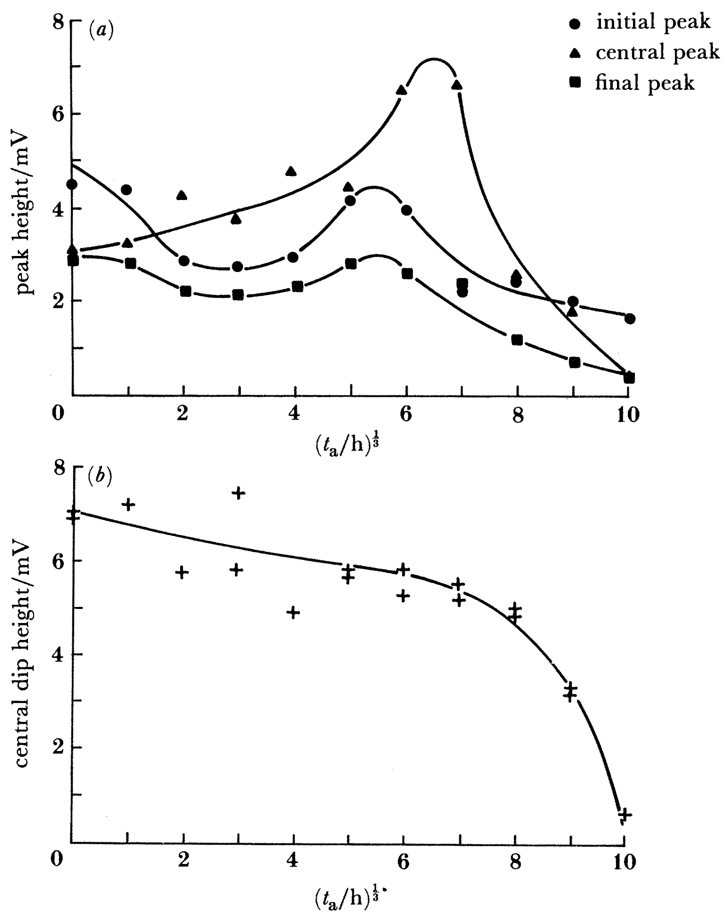

Figure 4 (*a*). Initial, central and final peak heights of BE activity and (*b*) central dip height of the MAE activity profiles as a function of $t_a^{\frac{1}{3}}$ for polycrystalline Incoloy 904 alloy. (The lines drawn on this figure and all the following figures are used merely to link points representing the same physical measurements.)

before falling away rapidly after longer times. The height of the initial peak at first falls after ageing and then rises to show a less prominent maximum than the central peak at ca. 170 h. The height of the final peal shows a similar trend.

The MAE activity profiles have just two peaks, which occur in the knee regions of the hysteresis loop as in figure 2*b*. In figure 4*b* the MAE central dip height is plotted as a function of $t_a^{\frac{1}{3}}$. Here the activity decreases slowly up to ca. 512 h and then more rapidly after longer ageing times. The initial and final peak heights show similar trends. For comparison figure 5 shows the mechanical hardness as a function of $t_a^{\frac{1}{3}}$, measured by using a 10 kg weight on a Vickers machine.

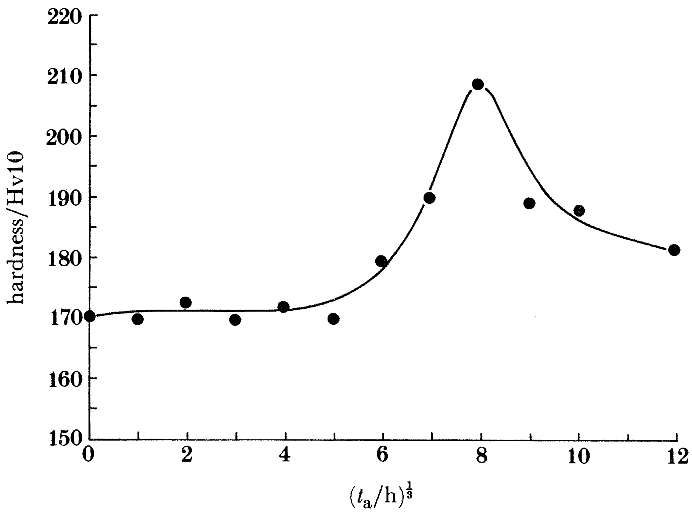

FIGURE 5. Vickers hardness as a function of $t_a^{\frac{1}{3}}$ for polycrystalline Incoloy 904 alloy.

The central peak, present in the BE activity profiles, is absent in the MAE activity profiles, indicating that the low field activity is predominantly due to irreversible 180 ° domain-wall motion. The peak in the BE activity at $t_a \approx 300$ h occurs where there is still an increase in the coercivity. It seems to occur where the inclusion diameter is similar to the domain-wall width, and closure domains form as observed in transmission electron microscopy studies (Taylor 1981).

The MAE central dip height does not depend on the inclusion diameter for $t_a \leqslant 343$ h. This implies that 90° domain walls are little affected by small inclusions. However, for longer ageing times (and hence larger inclusions), the activity falls off rapidly in a similar manner to the BE activity.

The peak in the mechanical hardness occurs at $t_a \approx 520$ h, indicating that slightly larger inclusions than in the case of magnetic peak ageing have the strongest influence on dislocations. Otherwise the trend is similar to that of the BE activity peak heights.

In practical testing, because of the observed trend in the peak heights of the MAE activity, provided that they are larger than the domain wall width (180 nm in Incoloy 904), it should be possible to use the technique to determine the precipitate size uniquely, whereas the BE activity can give ambiguous results. MAE would also give a more realistic result in materials with an inhomogeneous microstructre, as it is a bulk measuring technique.

If it were required to measure residual stress by BE or MAE in this material, then it would be necessary to take into account the effect of the inclusions, although this would not be necessary for MAE if their diameter were less than 100 nm.

4. Investigation of plastic deformation of pure α-Fe

The purpose of this experiment is to show how MAE and BE can be used to monitor plastic deformation and its removal by isochronal annealing in pure polycrystalline iron.

Four cylindrical tensile specimens of gauge length 45 mm and diameter 2.3 mm were machined from bars of pure iron (obtained from Johnson Matthey Chemicals Ltd., with an impurity concentration of 15 p.p.m. (by mass). They were annealed in vacuum to 950 °C for 16 h before plastically deformed by 4.9% at a strain rate of $1.85 \times 10^{-4}\ s^{-1}$ by using a 5 kN Instron testing machine. This introduced a reasonably high and homogeneous dislocation density (Leslie *et al.* 1963; Brandon & Nutting 1960). The ends were then removed by sparkcutting, reducing the specimen length to 40 mm. The specimens then underwent a series of vacuum anneals at successively higher temperatures for 1 h each and measurements were made before and after each anneal. The anneal temperatures varied from 250 °C to 750 °C in 100 K intervals.

After cold-work, the BE activity increased by about one order of magnitude over the whole field range. There were three peaks in the profile, one at low fields where $\mathrm{d}M/\mathrm{d}H$ is large and one each at the fields corresponding to the knees of the hysteresis loop (similar to aged polycrystalline Incoloy 904). These peak heights were subsequently measured after each anneal and are plotted in figure 6*a*. Similarly, figure 6*b* shows the peak heights from the MAE activity profiles for comparison. At low fields there is no peak in the MAE activity, indicating that the BE activity at low fields is predominantly, although not entirely, due to irreversible 180° domain-wall motion. However, measurements made on cold-worked nickel show a large peak in the MAE activity profile at low fields, implying that, in nickel, walls with $\theta \neq 180°$ interact strongly with dislocations, in contrast to iron. On annealing out the cold-work in nickel, the MAE activity profile reverts to the two peaks form as in iron. These observations on the relative strength of interaction of domain walls with dislocations are in agreement with calculations made by Scherpereel *et al.* (1970).

The main trend for both the BE and MAE activity is the large reduction in the peak heights after annealing at 650 °C. After this anneal the dislocation density would be expected to decrease dramatically (Keh 1962; Leslie *et al.* 1963). Measurements of dislocation density made by Keh (1962) after annealing at 550 °C indicate that the dislocation density is reduced by about 10–20% after 1 h, and that the dislocations would be straightened. The BE activity central peak height is reduced by 26%, whereas for MAE it is actually larger at 550 °C. This can be interpreted in terms of the stress relief caused by the straightening of dislocations. This would tend to increase MAE more than BE because 90 ° walls are more sensitive to stresses than 180 ° walls (the MAE in iron is reduced under stress (§6 and Kusanagi *et al.* 1979*a*)). A reduction in dislocation density, therefore, would reduce both MAE and BE but the accompanying reduction of stress would be expected to increase MAE again. The increase in the outer peak height at low temperatures is possibly due to the precipitation of a small number of carbides (Buttle *et al.* 1986).

Both MAE and BE are sensitive to dislocations and plastic deformation, and this suggests that

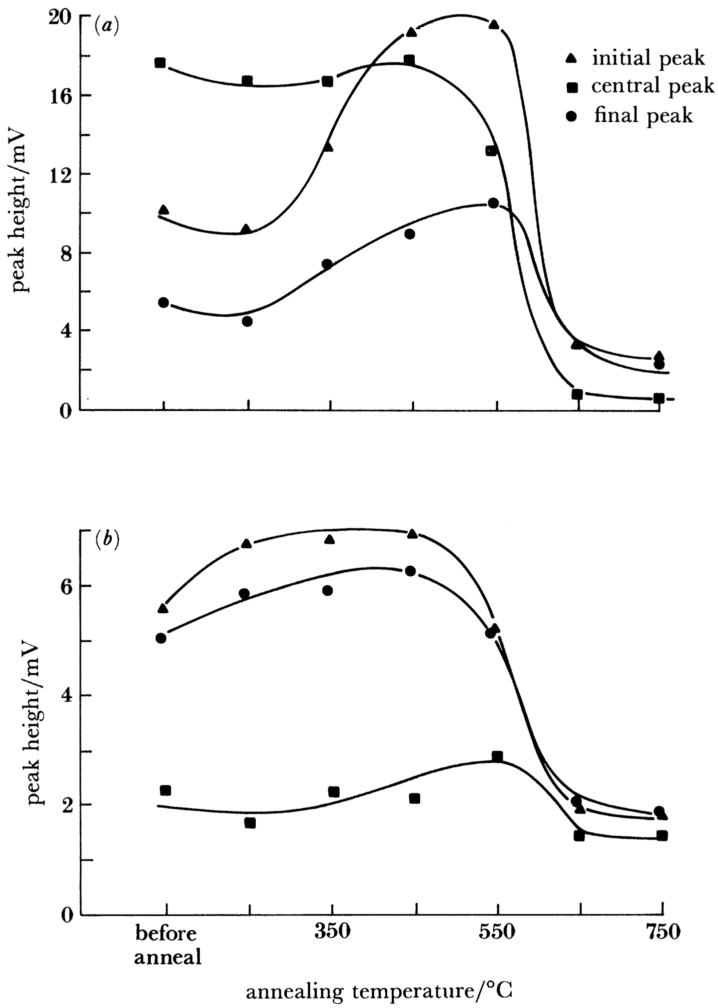

FIGURE 6. Initial, central and final peak heights of (*a*) BE and (*b*) MAE activity profiles as a function of annealing temperature for polycrystalline iron.

either technique could be used to monitor surface damage. It is clear that for residual stress measurement the effects of cold-work will have to be taken into account, and this may best be done by using both techniques, taking advantage of their different sensitivities by stress.

5. Investigation of radiation effects in an Fe–Cu alloy

These experiments were carried out to explore the potential of the MAE and BE techniques for monitoring the irradiation hardening and embrittlement in pressurized water reactor (PWR) pressure-vessel steels. These phenomena, which occur during 290 °C-neutron irradiation, are known to be associated with the presence of residual levels (*ca.* 0.1–0.35 %) of copper in steels. The present studies were therefore carried out on an Fe–0.2 % Cu alloy as a model system.

The role of copper in the radiation hardening or embrittlement process may be explained in terms of either: (*a*) segregation of copper to stabilize the point defect clusters (dislocation loops or microvoids) formed at the centre of displacement spikes by neutron–atom collisions or (*b*) precipitation strengthening by the FCC ε-Cu phase, formed as a result of the enhanced diffusion associated with the radiation-induced vacancy flux. The first mechanism is suggested by transmission electron microscope and field ion microscope observations (Smidt & Sprague 1973; Brenner *et al.* 1978). Arguments advanced for irradiation-induced copper precipitation (Odette 1983; Fisher *et al.* 1984) are based on (i) the strong age-hardening response of copper in α-Fe, together with (ii) the low equilibrium solubility of copper in α-Fe at 290 °C (less than 0.02 %) (Salje & Feller-Kniepmeier 1977).

Although the full details of these mechanisms are still under discussion, it is clear that the microstructural features induced by irradiation, either in the form of precipitates or defect clusters, act as strong obstacles to dislocations and give rise to the observed radiation-hardening response; furthermore, such obstacles may also result in domain-wall pinning, and thereby be capable of detection by the present techniques.

Specimens of the Fe–0.2 % Cu alloy were solution treated at 850 °C for 18 h in vacuum and water quenched to retain copper in solid solution. Three specimens were then neutron-irradiated at 290 °C for a period of 120 h to a dose of 2.4×10^{19} n cm^{-2} (greater than 1 MeV). Three control specimens were aged for 120 h at 290 °C to isolate purely thermal effects from those induced by irradiation. Finally, irradiated and thermal control samples were isochronally annealed for 1 h periods in 50 °C increments over the temperature range 400–650 °C. MAE and BE activity profiles were measured before and after irradiation, before and after thermal ageing, and following each intermediate heat treatment during isochronal annealing. Mechanical property changes were followed throughout by using room temperature Vickers hardness tests at 10 kg load.

Figure 7 illustrates the hardness changes induced by neutron irradiation of the Fe–0.2 % Cu alloy, together with the effects of 290 °C thermal ageing and isochronal annealing. These data demonstrate that irradiation produces marked strengthening with a hardness increase greater than 100 %, and that maximum recovery of the hardness change occurs over the range 550–650 °C. This behaviour is well established and can be explained in terms of the formation of point-defect clusters, copper precipitates or both, followed by their annealing out and overageing respectively. The control specimens exhibit a small hardening peak centred at 550–600 °C which suggests limited thermal formation and overageing of copper precipitates.

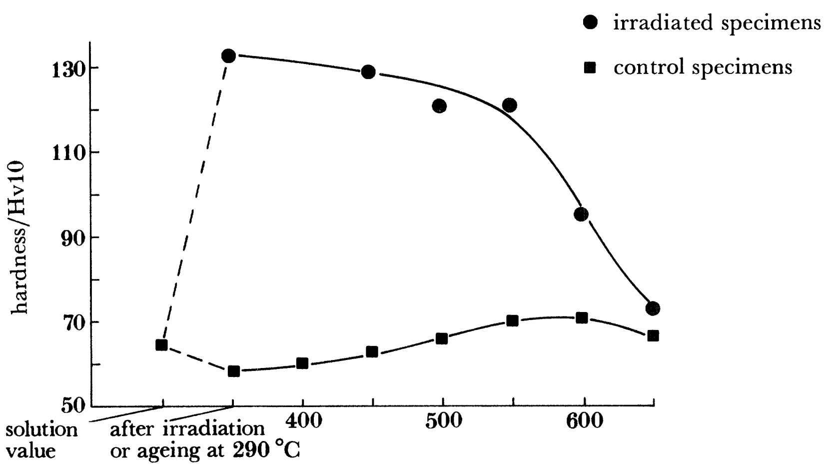

FIGURE 7. Effects of neutron irradiation, thermal ageing and isochronal annealing on room temperature hardness of Fe–0.2 % Cu alloy.

The corresponding trends in MAE and BE at low field with irradiation, ageing and isochronal annealing are illustrated in figure 8. Both thermally induced and irradiation-induced effects are observed. The increases, particularly in BE after thermal ageing alone, are large. One explanation is that the vacancy supersaturation quenched in at 850 °C induces copper precipitation. This vacancy concentration will be greatest at the surface of the specimen (subject to maximum quench rate) and this is reflected in the surface-sensitive BE activity. The decay of BE activity with isochronal annealing is a consequence of the re-solution or overageing of these precipitates.

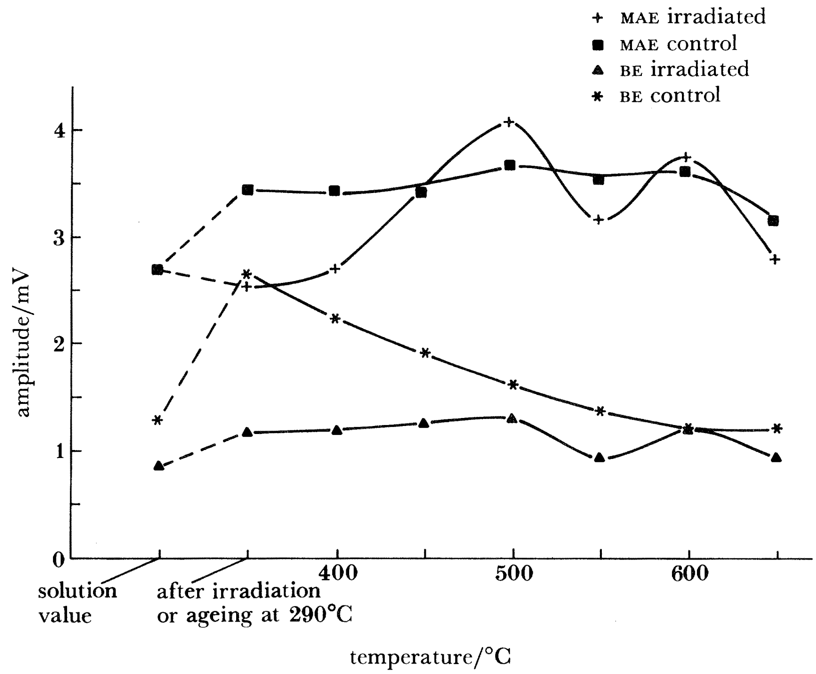

FIGURE 8. Effects of neutron irradiation, thermal ageing and isochronal annealing on MAE and BE activity profiles (central dip heights) of Fe–0.2 % Cu alloy.

The effect of neutron irradiation is to induce a small overall increase in BE, but little change in MAE, indicating increased pinning of 180 ° domain walls. However, the individual Barkhausen events are much smaller, indicating considerable enhancement of pinning density with irradiation.

The most significant changes, however, occur during post-irradiation annealing. Specifically, there is a steep increase in MAE over the temperature range 400–500 °C, together with well defined peaks at 500 and 600 °C; these changes are mirrored, albeit on a much reduced scale, in the BE activity. The monotonic increase in MAE up to a peak at 500 °C, followed by a sharp cut off, is analogous to the effects of cold-work (*cf* §4), implying a mechanism based on the annealing out of dislocation loops; the temperature range for the second peak (600 °C) is such that this would be consistent with the coarsening of Cu precipitates formed during irradiation. Clearly, the precipitate sizes responsible for peak MAE and BE activity do not correspond to that of peak strengthening in this material. Thus, precipitate coarsening at 600 °C leads to loss of hardening, but enhanced domain wall pinning as the particle diameter to domain wall width approaches unity; see §3.

In summary, the results of the MAE and BE studies imply that two independent irradiation strengthening mechanisms are operative in Fe–0.2 % Cu alloy after 290 °C-neutron irradiation, one based on dislocation loop formation and a second attributable to Cu precipitation. The MAE technique, in particular, appears to be capable of detecting and distinguishing between domain wall pinning caused by both obstacle types, but only during post-irradiation annealing; this is a consequence of the dependence of the sensitivity of the MAE activity on the size of defect relative to domain wall width. Correlations between MAE and BE, and other detection methods capable of characterizing matrix defects at sizes 1–2 nm (e.g. small-angle neutron scattering and field-ion microscopy) could be useful in confirming the above interpretation, and in providing a calibration of the technique for use in monitoring embrittlement levels in nuclear pressure vessels.

6. Investigation of tensile stress dependence

The purpose of this experiment was to monitor the effect of applied tensile stress on MAE and BE activity in iron, nickel, mild steel, and Incoloy 904.

For materials with a positive magnetostriction, the application of a tensile stress will cause the domains that are magnetized in directions closest to the stress direction to grow at the expense of neighbouring domains, in order to reduce the elastic energy. In the present experimental arrangement the magnetization direction is parallel to the stress axis. Consequently, as the tensile stress is increased in combination with a periodic magnetic field, there is a greater tendency for domains with their magnetization direction parallel to the stress axis, and therefore 180° domain wall movements, to become dominant. For materials with a negative magnetostriction the effects are more complex.

MAE and BE activity profiles were measured from specimens of iron, nickel, mild steel and Incoloy 904 as a function of applied tensile stress. Cold-worked iron, mild steel and Incoloy 904 all have a positive magnetostriction constant and so, from the above argument, we may expect that the application of a tensile stress will increase the BE low field activity (predominantly due to 180° domain-wall motion), and reduce the MAE low field activity (due to 90° domain-wall motion). This was observed to be the case, there being 7 %, 22 % and 40 % increases, respectively, in the BE central-peak height, and 50 %, 20 % and 5 % reductions, respectively, in the MAE central-dip height, on applying a stress of 23.6 MPa. In nickel, which has a negative magnetostriction constant, the BE activity central peak showed a reduction of 4 % at 23.6 MPa and 29 % at 94.4 MPa, whereas for the MAE there were also reductions of 16 % and 42 % respectively.

Figure 9 (*a*, *b*) shows the initial, central and final peak heights as measured from the BE and MAE activity profiles for cold-worked mild steel (with yield point of 356 MPa). In materials with a fairly large pinning density such as those of § 3 and 4 the BE activity profile is usually dominated by a large central peak at low fields and small ledges or peaks at higher fields (at the knee regions of the hysteresis loop). In practical testing, these outer regions of activity would be more difficult to investigate than the central peak region, although the central peak region is mainly due to 180 ° domain-wall movements and less sensitive to stress. In the case of mild steel (figure 9*a*) the sensitivity of the central peak height to stress falls off above 40 MPa. However, the MAE outer peaks (which tend to dominate the profile) show high sensitivity up to 140 MPa. Figure 10(*a*,*b*) shows the three parameters measured from the MAE activity profiles

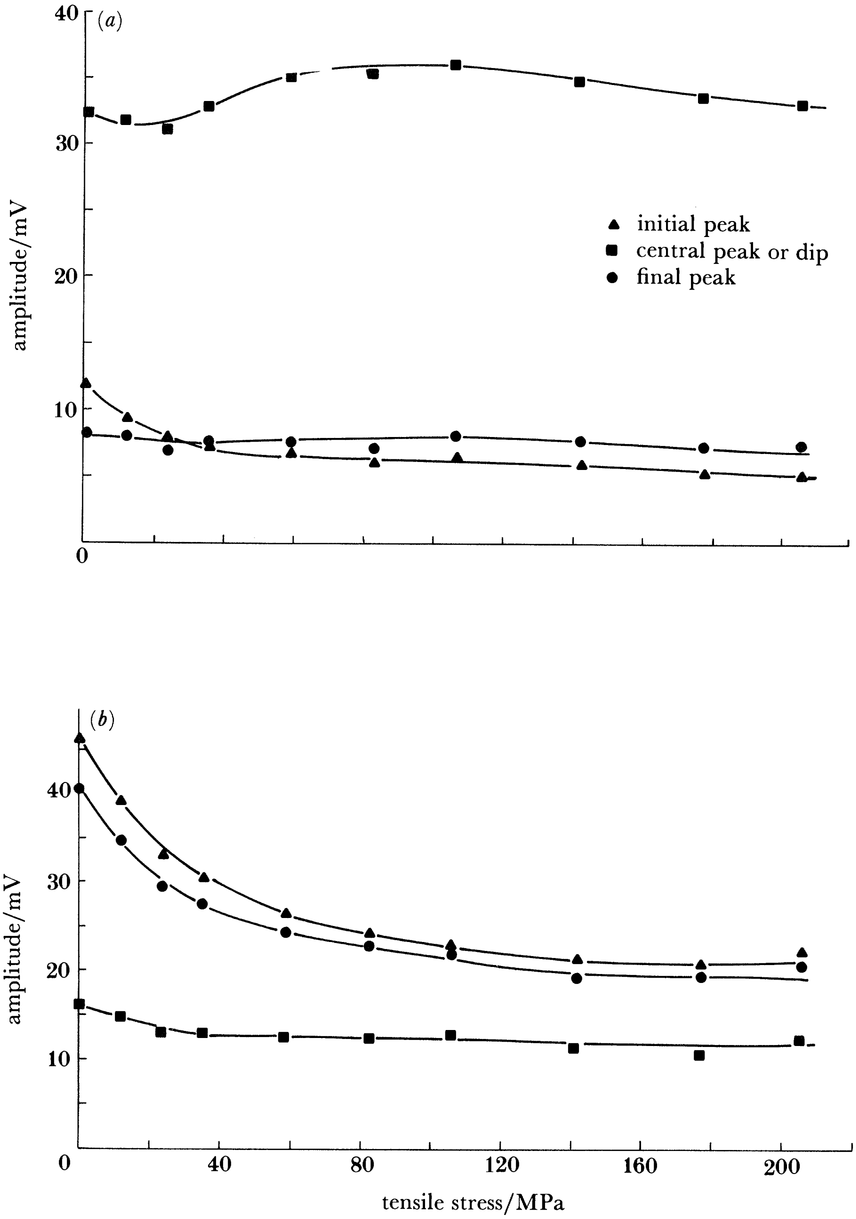

FIGURE 9. Initial, central and final peak heights of (*a*) BE and (*b*) MAE activity profiles as a function of tensile stress for mild steel.

in Incoloy 904, (*a*) being for the unaged material and (*b*) for material aged for 343 h at 700 °C. Again the MAE is sensitive to stress over the whole range plotted (100 MPa). For elastic strains BE tends to be sensitive over a fairly small range and is of course limited to a small surface layer (10–100 μm). Although the magnetization process used in this MAE experiment involves a 25 Hz frequency component, thereby limiting measurements to a *ca.* 0.1–1.0 mm surface layer, measurements can be made at lower frequencies (Buttle *et al.* 1986).

From figure 10 (*a*, *b*) it is apparent that the microstructure modifies the stress dependence

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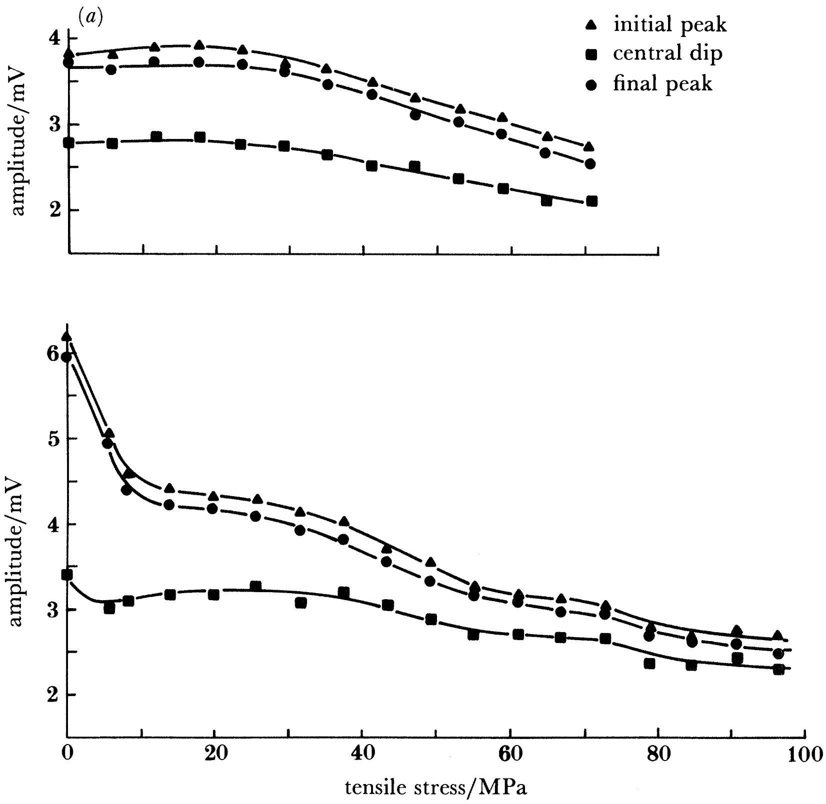

Figure 10. Initial, central and final peak heights of MAE activity profiles as a function of tensile stress for polycrystalline Incoloy 904 alloy; (*a*) unaged (*b*) aged for 343 h at 700 °C.

quite considerably, particularly for stresses less than 10 MPa, the aged material showing the strongest dependence. It is clear that further work is necessary, to be able to separate microstructural effects from stress effects.

7. Conclusions

The potential of MAE and BE for non-destructive examination have been assessed. These techniques show sensitivity in four areas: to precipitates in Incoloy 904, dislocations in α-Fe, tensile stress and radiation damage. Their strong sensitivity to precipitates and dislocations suggests that they could be used to monitor precipitate hardening and cold-work or plastic damage, respectively. The sensitivity to radiation damage is small but nevertheless useful for characterizing radiation effects. One of the most useful applications may be in the non-destructive measurement of stress, particularly residual stress. However, for practical monitoring of stressed regions in a structure, allowance may have to be made for possible accompanying variations in the local microstructure. Further measurements of the effects of microstructure and stress on MAE and BE are therefore recommended.

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Discussion

G. Busse (*I.K.P., Universität Stüttgart, F.R.G.*). What size of sample is needed? Is it possible to obtain spatially resolved information in terms of scanning along the rod axis?

D. J. Buttle. We used small cylinders of length 40 mm and diameter 2.3 mm in our experiments. There is no reason why the technique should not be used on different geometries provided that the material can be adequately magnetized, although in the case of the BE measurement a surface rather than an encircling coil would have to be used for larger structures. Other workers have made measurements on plates and sheets (see, for example, Kusanagi *et al.* 1979). The sample geometry does modify the MAE signal, however, since much of the acoustic energy goes into various resonant modes of the specimen.

There are two possibilities for obtaining spatially resolved information. The first would be to sweep a localized magnetizing field along the length of the rod, although this would only give a very crude resolution especially for short rods. The second approach would be to use a transducer array to locate the source of the larger MAE events; this approach has been successfully used at Harwell to locate acoustic sources from crack growth during tensile testing.

C. M. SAYERS (*Materials Physics and Metallurgy Division, AERE Harwell, Oxfordshire, U.K.*). How sensitive are Mr Buttle's results to the magnetic history of the material? For example, if he measured the amplitude against stress curves for two samples of mild steel which have been cooled at different rates in a magnetic field, how different would the results be?

D. J. BUTTLE. Leaving aside the different microstructures which may arise due to the different cooling rates (e.g. if they were cooled from around 1000 °C they would have different pearlite fractions and plate spacings etc., which would certainly give rise to different characteristic signals), in a soft magnetic material cooling in a magnetic field is unlikely to modify the microstructure except for a possible effect on the arrangement of point defects. This is unlikely to show up strongly in BE or MAE and would probably be masked by the difference in microstructure due simply to the difference in cooling rates. In some materials cooling in a magnetic field could introduce texture, but this is unlikely to happen in steels. The two specimens might have a different remanent magnetization, but any remanent field remaining in the material would not affect the results since the specimens are first taken through a complete magnetization cycle before measurements are made.

R. E. GREEN (*Center for Nondestructive Evaluation, The Johns Hopkins University, Baltimore, Maryland, U.S.A.*). This is not a question directed to the speaker, but a challenge to his co-authors at Oxford and Harwell. Because one of the major problems facing successful practical implementation of acoustic-emission techniques for non-destructive evaluation of materials is that the specific emissions from the various microstructural alterations causing the emissions is unknown, broadband transducers must be used to detect and record as large a frequency spectra of sounds from the test object as possible, Unfortunately, numerous sources of background noises are present in all practical working environments and many of these noise sources possess acoustical frequencies in the same range as the acoustic emission signals themselves. Therefore, the major problem preventing optimization of acoustic emission non-destructive testing is separation of a meaningful signal from the noise background.

Because Harwell has experts in both theory and experiment dealing with acoustic emissions and because Oxford has the most outstanding electron microscope facilities in the world, why no conduct some acoustic-emission experiments in these microscopes to definitively determine the specific acoustic emission signals associated with a given microstructural alteration? Such information would be extremely valuable.

D. J. BUTTLE. The reason that we have used model systems such as Incoloy and pure iron has been precisely to isolate different mechanisms of interaction of the domain walls with microstructure, and we have been able to draw extensively on dynamic transmission electron microscopy (TEM) observations of such interactions in interpreting our results. We agree about the usefulness of the kind of experiments suggested. The size of specimens for TEM is so much smaller than any that has been measured in MAE experiments that the difficulty in combining the techniques should not be underestimated. Scanning electron microscopy or X-ray tomography may prove more suitable for observing the domain walls while detecting MAE.

Regarding the problems facing the practical implementation of acoustic emission techniques to which Professor Green refers, we agree with his preference for broad-band transducers for source characterization. Although we plan to apply broad-band techniques to MAE in the future, we recognize that they may be limited to low-noise environments. This is because MAE signals tend to be smaller than those generated by for instance fast crack-growth, where source characterization has been successfully demonstrated. It is thus likely that relatively narrow-band detection will be necessary to ensure adequate sensitivity in many industrial applications. Work at IzfP, Saarbrucken, West Germany, for example, has shown that MAE and BE data can be correlated with materials properties in a more industrial environment.

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