The sources of acoustic emission in a steel of the A533B composition with low sulphur content have been investigated using two different recording techniques. First, systematic narrow band (0.1–1.0 MHz) measurements were made of acoustic emission activity during uniaxial tensile tests of specimens with a range of controlled microstructures. Mechanical properties and fractographic data were analysed, and the most probable deformation and fracture mechanisms deduced for these microstructures. It appears that only a few processes generate detectable emission in this material. For instance, high-velocity dislocation motion can be detected during the yield of slow-cooled or long-tempered specimens, and cleavage fracture in rapidly quenched specimens. A model is presented to relate the magnitude of the emission to the dynamic operation of dislocation and fracture sources. Second, very wide band (25 MHz) calibrated measurements were made of the most energetic emission sources in rapidly quenched material. These indicate the sources to be consistent with cleavage microcracking. A typical microcrack propagated over a distance of \( \sim 50 \mu m \) with an average velocity of \( 450 \text{ m s}^{-1} \).

MS/0627


When a sensitive piezoelectric transducer is attached to a metal undergoing deformation, motion of the surface can often be detected. This motion is very small in amplitude, with displacements of \( 10^{-14} \text{ m} \) or less, and can also be very rapid, with frequencies extending to 50 MHz or more. \(^1\) This surface motion is called acoustic emission and is believed to be generated by localized relaxations in the elastic strain distribution within the metal. \(^2\) These relaxations are usually short lived and occur, for instance, when dislocations or cracks rapidly propagate. It is probable that every mobile dislocation and every growing crack will cause the generation of elastic waves that propagate to the surface of metal. However, due to the limited sensitivity and frequency response of piezoelectric transducers (typically \( \sim 10^{-14} \text{ m} \) and \( \leq 1 \text{ MHz} \)) only some of the deformation or fracture processes that occur are detected \(^1\) in most acoustic emission tests.

Thus, the acoustic emission technique can only be used to investigate deformation and fracture processes which generate surface displacements greater than the detection threshold, and whose frequency content is at least partially covered by the range of the detection instrumentation. These two criteria for detectability are partly responsible for the strong dependence of acoustic emission activity upon testing variables (sample geometry, strain, strain rate, temperature) and metallurgical variables (crystal structure, composition, grain size, precipitate structures, etc.). While there is now a reasonable empirical understanding of some of the effects of test variables such as strain rate or gauge volume, the effects of metallurgical variables are much less clearly understood, especially in ferritic steels where the occurrence of the austenite to ferrite transformation on cooling from above ~850°C results in complex microstructures. \(^2\)

From the practical viewpoint, when acoustic emission is used to detect growing flaws, it is most important that one should have a good understanding of the origins of the detectable acoustic emissions from ferritic steels. Previous studies have indicated that some important types of crack-growth process, e.g. ductile fracture by void growth and coalescence, may not be detectable in all ferritic steels. \(^3\) In these materials one must look for acoustic emission from plastic zone deformation processes in order to detect the presence of growing flaws.

Detailed metallurgical studies of the sources of acoustic emission during plastic deformation have indicated that in ferritic steels the fracture/decohesion of inclusions \(^4\) \(^5\) (and possibly carbide precipitates) together with the sudden motion of groups of dislocations in the vicinity of general yield \(^5\) are detectable under some conditions. There is also good evidence that acoustic emission is generated during brittle fracture below the ductile to brittle transition temperature or following temper or hydrogen embrittlement. \(^6\) Clark and Knott \(^7\) have suggested that at least one ductile mechanism of fracture, that of alternating shear fracture, may also be a source of energetic emission. Both the occurrence of these emission sources and the properties of the emission they generate may change in a steel of fixed composition but which undergoes different thermomechanical treatments before testing. \(^5\)

In this work tensile specimens were used to study systematically the effect of changes in microstructure of an experimental ferritic steel of fixed composition upon acoustic emission activity. The composition chosen for study was 0.2%C, 1.3%Mn, 0.5%Mo, 0.5%Ni and was similar to that of A533B. The sulphur content was reduced to a minimum and the forging treatments adjusted to avoid large numbers of elongated MnS inclusion stringers, a prominent acoustic emission source in some experiments. Thus, the experiments were designed to determine the acoustic emission from metal deformation processes in the absence of emission from elongated MnS inclusions.
Table 1 Bulk chemical composition of Mn–Mo–Ni alloy, wt-%

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>S</th>
<th>Mo</th>
<th>Ni</th>
<th>Cu</th>
<th>N</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>Content</td>
<td>0·24</td>
<td>0·27</td>
<td>1·35</td>
<td>0·0045</td>
<td>0·5</td>
<td>0·55</td>
<td>0·02</td>
<td>0·001</td>
<td>0·005</td>
</tr>
</tbody>
</table>

The majority of tests in the study were performed using conventional 'narrow band width' measurements with piezoelectric transducers. These measurements are inherently limited and provide only qualitative information about the acoustic emission source. To overcome the limitations of the narrow band technique, additional tests were performed with calibrated measurements up to 25 MHz, using capacitance transducers, in order to characterize the most energetic emission sources.

EXPERIMENTAL

Material
A 100 kg vacuum-degassed ingot of a Mn–Mo–Ni steel of A533B nominal composition was produced by Ross and Catherall Ltd using a low-sulphur Japanese Electrolytic iron. The bulk chemical composition of the ingot was determined at three positions across the cross-section for each of five equally spaced locations along the ingot length. These 15 analyses showed there was no significant macrosegregation. The mean concentration values for detected elements are given in Table 1. Chemical analysis indicated the phosphorus concentration was less than 200 ppm, the trace impurities Sn, Sb, and As were less than 100 ppm, while Ti, V, Nb, and Cr were all less than 40 ppm in concentration.

The ingot, originally ~60 cm long and ~16 × 16 cm in cross-section, was cut into four shorter billets each ~13 cm long. These were heated at 1050°C for 2 h and press forged to 5 cm thick plate so that the plane of the final plate contained the ingot solidification direction.

Small samples of the forged plate were examined for evidence of microsegregation. The metal samples were polished by standard metallographic techniques, lightly etched in 2% nital, and the surface composition measured using a Cameca X-ray microprobe analyser. Using an electron beam with a spot size of ~1 μm, significant variations in Mo, Mn, and Ni concentrations were detected (Fig. 1). Molybdenum showed the severest segregation; some regions contained up to four times the mean Mo concentration. The segregation in two dimensions of the molybdenum is shown in Fig. 2. The spacings between the Mo-rich areas varied with orientation in part as a consequence of the non-uniform forging deformations; on average, the spacing was ~0·5 mm. Silicon and manganese cosegregated with the molybdenum. Attempts were made to reduce the segregation by prolonged austenitizing treatments at temperatures above 1100°C. These were successful at reducing the segregation to insignificant levels but resulted in an unacceptably large prior austenite grain size. Consequently, for the studies reported here, some alloy-element segregation was present.

Specimen geometries
Two specimen geometries were used. For the majority of tests where narrow band measurements of relative acoustic emission activity were made, a modified dumb bell specimen was used (Fig. 3a). The tensile axis of the specimen was parallel to the solidification direction of the original ingot. The gauge was 2 mm long × 3 mm dia. and this short gauge section was used in order both to simplify interpretation by localizing the emission source and to facilitate comparison with measurements of the Yobell specimen (Fig. 3b) which had been developed for wide band measurements of the acoustic emission wave form. It was used here to characterize the more energetic emission sources found from narrow band experiments.

Heat treatments
Specimens were given one of two types of heat treatment. In one, dumb bell and Yobell specimens were austenitized 1 h at 1000°C in a dynamic argon atmosphere and cooled at varying rates. The differing cooling rates were achieved by quenching different specimens into either iced brine at ~10°C, 10% NaOH solution at room temperature, unstirred water, oil, or by air cooling or furnace cooling at ~50 K h⁻¹. These treatments resulted in a wide range of

1 Relative concentration profiles for Si, Ni, Mn, and Mo in hot-forged material; major alloying elements show differing degrees of microsegregation

2 Segregation of Mo in Mn–Ni–Mo steel after hot forging; darkest shading corresponds to highest concentration of Mo
cooling rates with a correspondingly large range of microstructures, mechanical properties, and acoustic emission activities. There was no evidence of quench cracking in any of the specimens.

The second set of treatments were performed on dumb bell specimens only. Specimens were water quenched from 1000°C and pairs subjected to isothermal tempering treatments at 650±5°C in argon for times of 6-10 020 min, and water quenched. It was recognized that those specimens tempered for the shorter times had probably not reached 650°C for more than a few minutes before quenching. All specimens were then electropolished in a 10% solution of perchloric acid in acetic acid to remove the surface oxide layer as the cracking of this layer has been found to be a copious source of extraneous acoustic emission.

**Acoustic emission measurements**

Acoustic emission measurements were made with two different recording systems. First, a more conventional narrow band technique incorporating a piezoelectric transducer was used to determine systematically the relative emission activity of the dumb bell specimens. Second, a wide band calibrated technique incorporating a capacitance transducer and digital recording was used in an attempt to characterize the emission sources in Yobell specimens.

**Narrow band technique**

Dumb bell specimens were tested in tension using a model 1195 Instron screw-driven machine. Specimens were gripped at the 45° shoulders at either end of the specimen. All tests were performed at a constant crosshead displacement rate of 0.2 mm min⁻¹ (~1.7 x 10⁻³ s⁻¹ strain rate). The grips contained special insulators which were
used to isolate electrically the specimen from the testing machine in order to reduce ground-loop coupled noise. However, these insulators were mechanically soft and thus a component of the crosshead displacement was accommodated in the grips rather than the specimen gauge. During each test the load was continuously monitored together with the acoustic emission activity.

The acoustic emission system is shown schematically in Fig. 4a. Care was taken to standardize all experimental procedures so that valid comparisons could be made between different tests. The acoustic emission was detected with a damped piezoelectric transducer attached by a thin layer of grease, under a constant small pressure, to the polished end of each specimen. The transducer had an approximately flat frequency response below 1 MHz and was connected to a wide band preamplifier which both amplified and differentiated the transducer output. Thus, the output voltage of the head amplifier was expected to be proportional to the surface velocity of the specimen. The sensitivity of the system to small-amplitude emissions was improved by reducing the ambient noise with acoustic and radio frequency shielding. A 0.01-1.0 MHz bandpass filter was used to increase signal/noise ratio. The signal was further amplified and monitored with a Hewlett-Packard 435A power meter which permitted the electrical power due to an emission to be recorded. This was approximately proportional to the acoustic power impinging on the transducer in the pass band 0.1-10 MHz. The power-meter output was also integrated so that a measure of the energy of an emission could be obtained. Both the emission power and energy together with the load were plotted on a chart recorder as a function of crosshead displacement during a tensile test. The electrical signal was also continuously monitored on an oscilloscope.

**Wide band technique**

The objective here was to measure accurately the surface displacement wave forms from acoustic emission events during the deformation and fracture of Yobell specimens. While such wave forms can be measured for any specimen geometry, a meaningful analysis can be performed only if the effects of specimen boundaries are removed, or at least minimized. The Yobell geometry permits this while still allowing a simple mechanical test to be performed. All the Yobell specimens were tested to failure in an Instron 1195 screw-driven machine at a crosshead speed of 0.1 mm min\(^{-1}\). The transient displacement at the centre of the polished surface of the Yobell specimen (designed to be the epicentre), due to the arrival of an acoustic emission signal, was measured with a capacitance transducer (Fig. 4b). The transducer air gap was maintained at 2-7 \(\mu\)m for every test, giving a constant transducer sensitivity of 7.2 \(\times\) 10\(^{-3}\) C m\(^{-1}\).

The transducer was coupled to a charge-sensitive preamplifier with a short a length of cable as possible (about 20 cm in practice). This was followed by two filters: a high-pass filter at 30 kHz to eliminate low-frequency machine noise and a low-pass filter at 45 MHz to eliminate signal aliasing errors during later analogue to digital conversion. The filtered signal was further amplified before digitization in a Biomation transient recorder and storage in a PDP8/E minicomputer. The recording system had a flat transfer function of 5.35 \(\times\) 10\(^{3}\) V m\(^{-1}\) from 80 kHz to 25 MHz, the upper band width limit of the Biomation recorder. The recorder threshold was set just above the ambient noise and corresponded to a surface displacement of 1 \(\mu\)m.

**Mechanical properties data**

The low compliance of the Instron prevented accurate stress–strain measurements of the specimens tested for acoustic emission response. To overcome this, additional specimens of identical geometry and heat treatment were tested in a servohydraulic machine at a constant strain rate of \(\sim 2.1 \times 10^{-4}\) s\(^{-1}\). Knife-edge extensometers were mounted at the ends of the gauge length so that the strain within the gauge could be measured as a function of load throughout the test. After testing, the load–extension data were converted to give nominal stress–nominal strain and true strain–true plastic strain data, so that proof and ultimate stresses, strains to fracture, and work-hardening exponents could be measured.

**Microstructural characterization and fractography**

After testing, the undeformed ends of dumb bell specimens were ground and polished to allow the microstructure to be characterized by optical metallography. In addition, extraction replicas were taken from the polished and etched surfaces and the precipitate type and composition for some of the long-tempered specimens determined using microdiffraction techniques and energy-dispersive X-ray analysis. The precipitate size distribution was determined from scanning electron micrographs of nital-etched specimens.

The fracture surface of one specimen in each condition was examined by scanning electron microscopy to determine the fracture mode. In addition, fractured ends of specimens were also mounted, ground, and polished so that the longitudinal section behind the fracture face could be examined by optical and scanning electron microscopy.

**RESULTS**

**Characterization of microstructure**

All specimens were austenitized for 1 h at 1000°C which resulted in a prior austenite grain size of \(\sim 70 \mu\)m. Variation of cooling rate from 1000°C led to specimens of widely differing microstructures (Fig. 5). Both the iced-brine and NaOH quenches produced acicular martensite containing no visible carbides when examined by optical microscopy (Fig. 5a and b). Oil quenching (Fig. 5c), and to a lesser extent water quenching, also resulted in a martensitic structure which now contained barely resolvable carbides. Air cooling resulted in a bainitic microstructure (Fig. 5d), the visible carbides being \(< 0.5 \mu\)m in size. The much slower furnace-cooling treatment gave rise to a complex microstructure with randomly distributed areas of upper bainite, fine pearlite, and proeutectoid ferrite containing isolated carbides (Fig. 5e).

Isothermal tempering at 650°C of water-quenched specimens also resulted in major changes of microstructure (Fig. 6). Tempering for as little as 6 min (Fig. 6a) resulted in the appearance of small carbides on lath and prior austenite boundaries. Increasing the tempering time caused the growth of carbides (Fig. 6b–e). After tempering for about 4–8 h the largest carbides were \(\sim 1 \mu\)m in size and there was evidence of the formation of subgrains of ferrite, formed by recovery of the quenched-in dislocation structure. Further prolonged tempering increased the carbide size and led to the complete disappearance of martensite laths. These carbides are shown in Fig. 7. Carbides less than \(\sim 0.2 \mu\)m were unresolvable, which accounts for the fall in distribution at small carbide size. After testing, large carbides located close to the fracture surface had themselves cracked (Fig. 8).

Specimens tempered for more than 4 h were also
examined using convergent beam microdiffraction techniques. The majority of carbides present in these specimens tempered \( \leq 8 \) h were found to be of the M\(_3\)C type. However, prolonged tempering resulted in the appearance of M\(_{23}\)C\(_6\) precipitates, especially at prior austenite grain boundaries (Fig. 9). Energy-dispersive X-ray analysis revealed that the M\(_3\)C carbides had a Mn/Fe ratio of 0.15 \( \pm \) 0.03 and a Mo/Fe ratio of 0.04. The Mn/Fe ratio for M\(_{23}\)C\(_6\) was 0.13 \( \pm \) 0.03, similar to that of M\(_3\)C, whereas the Mo/Fe ratio was significantly greater at 0.08.

It was noted in the specimens tempered for the longest times that the apparent volume fraction of carbides varied over the specimen surface. This was thought to be a consequence of the diffusion of carbon to the micro-segregated alloying elements resulting in a non-uniform carbide distribution.
Mechanical properties
The effect of varying the quench rate upon the relationship between nominal stress and nominal strain is shown in Fig. 10. True stress $\sigma_T$ was also calculated as a function of the true plastic strain $\varepsilon_{TP}$ using the relations

$$\sigma_T = \sigma_n (1 + \varepsilon_n) \quad \cdots \cdots \cdots \cdots \quad (1)$$

$$\varepsilon_{TP} = \ln (1 + \varepsilon_{np}) \quad \cdots \cdots \cdots \cdots \quad (2)$$

where $\sigma_n$ is the nominal stress, $\varepsilon_n$ the nominal strain, and $\varepsilon_{np}$ the nominal plastic strain defined as

$$\varepsilon_{np} = \varepsilon_n - \sigma_T/E \quad \cdots \cdots \cdots \cdots \quad (3)$$

where $E$ is Young's modulus. The true stress v. true plastic strain data were plotted logarithmically to calculate the work-hardening exponent $n$ for each material condition.
Example of cracked carbide precipitate close to fracture surface of specimen tempered 167 h (Fig. 11). A summary of mechanical properties data and $n$-value estimates is given in Table 2 for each condition. As the cooling rate decreased on going from an iced-brine quench to a furnace cool, the 0.1% proof and maximum stresses decreased while the plastic strain to fracture and the apparent $n$ value showed a small increase. The true plastic strain at maximum nominal stress did not vary systematically with cooling rate.

The effects of isothermal tempering at 650°C for periods of 6–10 020 min upon mechanical properties are shown in Figs. 12 and 13 and summarized in Table 3. Tempering reduced the 0.1% proof and maximum stresses while causing an increase in plastic strains to maximum stress and to fracture and an increase in $n$ to 0.22.

**Fracture appearance**

Variation of cooling rate had a major effect upon the mode of fracture at room temperature. The iced-brine and NaOH-quenched specimens underwent a predominantly brittle fracture (Fig. 14a) even though preceded by significant plastic deformation (Table 2). The fracture surface could be divided into two regions: a central region that had fractured approximately transverse to the tensile axis and an annular outer region where a shear lip had formed. The central area fractured by transgranular cleavage (Fig. 14b), the individual microcracks appearing to have changed orientation at lath boundaries. Water-quenched specimens (Fig. 14c) exhibited only a few isolated areas of quasicleavage, the majority of the central region of the fracture had occurred by alternating shear (Fig. 14d).

Optical metallography of material adjacent to the fracture indicated that at the intersection of alternating shear facets a

**Table 2 Mechanical properties of specimens cooled at varying rates**

<table>
<thead>
<tr>
<th>Heat treatment</th>
<th>Nominal stress, MN m$^{-2}$ $\sigma_{0.5}$ $\sigma_{\text{max}}$</th>
<th>True maximum stress, MN m$^{-2}$ $\sigma_{\text{max}}$</th>
<th>True plastic strain at maximum stress $\sigma_{\text{max}}$</th>
<th>Nominal plastic strain to fracture</th>
<th>$n$ value</th>
</tr>
</thead>
<tbody>
<tr>
<td>1000°C, 1 h, iced-brine quench</td>
<td>1100 1790</td>
<td>2000 0.068</td>
<td>0.30</td>
<td>0.05</td>
<td></td>
</tr>
<tr>
<td>1000°C, 1 h, NaOH quench</td>
<td>1020 1690</td>
<td>1880 0.065</td>
<td>0.32</td>
<td>0.08</td>
<td></td>
</tr>
<tr>
<td>1000°C, 1 h, water quench</td>
<td>1110 1750</td>
<td>1945 0.068</td>
<td>0.32</td>
<td>0.01–0.19*</td>
<td></td>
</tr>
<tr>
<td>1000°C, 1 h, oil quench</td>
<td>905 1460</td>
<td>1580 0.057</td>
<td>0.33</td>
<td>0.04–0.15*</td>
<td></td>
</tr>
<tr>
<td>1000°C, 1 h, air cool</td>
<td>535 960</td>
<td>1035 0.061</td>
<td>0.51</td>
<td>0.08–0.32*</td>
<td></td>
</tr>
<tr>
<td>1000°C, 1 h, furnace cool 50 K h$^{-1}$</td>
<td>540 750</td>
<td>850 0.074</td>
<td>0.46</td>
<td>0.10</td>
<td></td>
</tr>
</tbody>
</table>

* $n$-value variation with strain.
Use of convergent beam microdiffraction and X-ray energy analysis to characterize carbide types and compositions in Mn–Mo–Ni steel of A533B composition.
crack occasionally propagated parallel to the tensile axis. Slower cooling rates resulted in the occurrence of a classical cup and cone fracture mode (Fig. 14e). The fracture surface was covered in fine dimples which were deepest in the slowest cooled specimen. In the furnace-cooled condition, spacing and size of dimples varied over the fracture. Some particles, comparable in size with the carbides observed by optical metallography, were observed within the dimples. This suggests that the large carbides acted as void-nucleation sites.

Tempering specimens, even for only 6 min, changed the fracture mode of water-quenched samples (Fig. 15a). The areas of alternating shear had almost fully disappeared and the fracture surface was covered in a fine array of barely resolvable shallow dimples (Fig. 15b). Increasing the tempering time had two effects. First, as can be seen by comparing Fig. 15a, c, and e, it led to the formation of a classical cup and cone fracture. Second, the size and depth of the fine dimples increased (Fig. 15b, d, and f). Few of the dimples in any tempered specimen contained resolvable particles. Some of those observed were spherical inclusions (Fig. 15f).

**Acoustic emission measurements**

**Narrow band tests**

Figure 16 shows the effect of quenching rate upon the stress and strain dependence of the acoustic emission from quenched specimens. The two most rapidly quenched specimens (iced brine and NaOH) behaved similarly: little or no emission was observed during elastic loading and only a few emissions were detected during post-yield deformation before attainment of maximum stress (Fig. 16a). As the specimens were deformed past maximum stress large-energy
burst emissions were generated and these increased both in magnitude and rate of occurrence as the specimen was strained towards final fracture. Similar behaviour was observed with water-quenched specimens (Fig. 16b) although in this case the onset of burst emissions was delayed until the final stages of deformation just before final failure. The oil-quenched (Fig. 16c) air- and furnace-cooled specimens generated little emission at any stage of deformation. The data for these tests have been summarized by measuring the acoustic emission energy for each test and are given in Table 4.

Tempering at 650°C also had a marked effect upon the stress and strain dependence of the acoustic emission power. Tempering water-quenched material for 6 min resulted in the complete disappearance of detectable emission except for a single emission at fracture (Fig. 17a). Tempering for longer times up to ~4 h gave similar results. However, specimens tempered for 8 h and more began to exhibit an acoustic emission peak in the vicinity of yield (Fig. 17b and c). This peak consisted of many small-amplitude overlapping signals (usually called continuous emission) with occasional larger amplitude signals superimposed. The acoustic emission energy data for these specimens are summarized in Table 5.

In all the tempered specimens the only emission generated after \( \sigma_{\text{max}} \) occurred at the instant of final fracture and this decreased with specimen tensile strength. The trends in emission activity are presented graphically in Fig. 18.

**Table 3 Mechanical properties of isothermally tempered specimens**

<table>
<thead>
<tr>
<th>Tempering time, min</th>
<th>Nominal stress, MN m(^{-2})</th>
<th>True maximum stress, MN m(^{-2})</th>
<th>True plastic strain to maximum stress</th>
<th>Nominal plastic strain to fracture</th>
<th>n value</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>1110</td>
<td>1750</td>
<td>1945</td>
<td>0.32</td>
<td>0.061</td>
</tr>
<tr>
<td>6</td>
<td>830</td>
<td>960</td>
<td>1066</td>
<td>0.46</td>
<td>0.061</td>
</tr>
<tr>
<td>13</td>
<td>635</td>
<td>940</td>
<td>1053</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>36</td>
<td>680</td>
<td>930</td>
<td>1045</td>
<td>0.47</td>
<td>0.07</td>
</tr>
<tr>
<td>60</td>
<td>680</td>
<td>870</td>
<td>971</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>120</td>
<td>700</td>
<td>850</td>
<td>957</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>240</td>
<td>695</td>
<td>795</td>
<td>906</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>480</td>
<td>680</td>
<td>785</td>
<td>870</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>1440</td>
<td>405</td>
<td>600</td>
<td>721</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>6000</td>
<td>270</td>
<td>510</td>
<td>642</td>
<td>0.47</td>
<td>0.068</td>
</tr>
<tr>
<td>10020</td>
<td>330</td>
<td>545</td>
<td>690</td>
<td>0.47</td>
<td>0.068</td>
</tr>
</tbody>
</table>

* n-value variation with strain.
14 Effect of quench rate upon fracture mode

Wide band tests
Five Yobell specimens were tested at an extension rate of 0.1 mm min$^{-1}$ using the screw-driven Instron machine. The specimens had been austenitized 1 h at 1000°C; two were quenched in iced-brine solution at $-8^\circ$C and the rest in water at room temperature. A representative stress vs. strain curve with superimposed acoustic emission activity is given for each condition in Fig. 19.

Acoustic emissions were detected by the capacitance transducer from all the specimens. These emissions, in the main, were generated after the attainment of maximum stress. The strain dependence of the emission activity of the Yobell specimens was qualitatively similar to that of dumb bells of similar quench method. More emissions were detected from the more rapidly cooled iced-brine quenched samples (see Table 6).
The calibrated detection system enabled each emission to be recorded as a time-varying surface displacement and one example from each condition is shown in Fig. 20. All of these surface-displacement wave forms had a characteristic pulse at the longitudinal (L) wave arrival time followed by a less distinct step at the arrival of the shear (S) wave. When account is taken of the low-frequency filtering of the measured wave forms, good agreement exists with the wave forms calculated for a modelled crack-growth event. The model did not predict the oscillatory form of the so-called ‘wash’ following the L arrival and preceding the S arrival. This may have been due to interactions of the generated wave with nearby free surfaces such as subcritical microcracks or the boundaries of the gauge.

Because of the generally good agreement of measured wave forms with theory, it was possible to apply the same analytical procedure as reported elsewhere\(^9\) to obtain the source function from each measured wave form. First, the
Scruby et al. Microstructure and acoustic emission in Mn-Mo-Ni steel

16 Strain dependence of stress and acoustic emission power for A533B steel austenitized 1 h at 1000°C

Table 4 Effect of quench rate on acoustic emission energy

<table>
<thead>
<tr>
<th>Quench method</th>
<th>$\leq \sigma_{\text{max}}$</th>
<th>$\geq \sigma_{\text{max}}$</th>
<th>Total</th>
</tr>
</thead>
<tbody>
<tr>
<td>Iced-brine quench</td>
<td>0.1</td>
<td>5.0</td>
<td>5.1</td>
</tr>
<tr>
<td>NaOH quench</td>
<td>0.6</td>
<td>4.5</td>
<td>5.1</td>
</tr>
<tr>
<td>Water quench*</td>
<td>0.3</td>
<td>2.5</td>
<td>2.8</td>
</tr>
<tr>
<td>Oil quench</td>
<td>0.0</td>
<td>0.7</td>
<td>0.0</td>
</tr>
<tr>
<td>Air cooled</td>
<td>0.5</td>
<td>1.1</td>
<td>1.6</td>
</tr>
<tr>
<td>Furnace cooled 50 K h$^{-1}$</td>
<td>0.3</td>
<td>0.3</td>
<td>0.6</td>
</tr>
</tbody>
</table>

* Average of four tests.

17 Strain dependence of stress and acoustic emission power for A533B composition steel water quenched and tempered at 650°C for various times

transfer function from source to detector was calculated. Second, the inverse transfer function was calculated which was convolved with the measured waveform to obtain the actual source function. This source function is in the form of a time-varying crack volume when the epicentre surface displacement is measured. The time dependence of the source function for the two representative waveforms shown earlier is also presented in Fig. 20.

In every case the volume rose to a positive maximum and then started to fall. The fall was artificial and was due to low-frequency filtering and for short-duration events this could be ignored. Previous work has shown that the two most useful parameters to measure are the maximum volume attained and the time taken. In order to reduce possible errors from gauge-boundary reflections the maximum volume was calculated as twice the volume reached when the surface displacement of the L arrival pulse reached its maximum value. This is equivalent to assuming that the true L wave pulse was symmetric about a vertical line drawn through its maximum. The time for which the source operated, the source lifetime, was likewise taken as twice that for the surface displacement at the L wave to go from zero to its maximum. The mean values of this maximum volume and lifetime are summarized for each
18 Acoustic emission generated $a$ before and $b$ after maximum stress as function of tempering time

19 Nominal stress as function of displacement for Yobell specimen tested in $a$ iced-brine quenched condition Y1 and $b$ water-quenched condition Y5; superimposed on each is acoustic emission activity; see text for discussion of parameter representing this activity, the source volume

20 Examples of acoustic emission displacement wave forms for $a$ iced-brine quenched specimen and $b$ water-quenched specimen; corresponding source-volume-time histories were calculated by convolution with inverse transfer function for specimen; apparent fall in volume at long times is due to high-pass filtering

Table 5 Effect of isothermal tempering on acoustic emission energy

<table>
<thead>
<tr>
<th>Tempering time, min</th>
<th>$\leq \sigma_{\text{max}}$</th>
<th>$\geq \sigma_{\text{max}}$</th>
<th>Total</th>
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<tbody>
<tr>
<td>0*</td>
<td>0.3</td>
<td>23</td>
<td>2.6</td>
</tr>
<tr>
<td>6</td>
<td>0.0</td>
<td>0.9</td>
<td>0.9</td>
</tr>
<tr>
<td>18</td>
<td>0.3</td>
<td>0.9</td>
<td>1.2</td>
</tr>
<tr>
<td>36</td>
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<td>0.9</td>
</tr>
<tr>
<td>60</td>
<td>0.1</td>
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<td>0.8</td>
</tr>
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<tr>
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</tr>
<tr>
<td>1440</td>
<td>1.9</td>
<td>0.4</td>
<td>2.3</td>
</tr>
<tr>
<td>6000</td>
<td>1.6</td>
<td>0.3</td>
<td>1.9</td>
</tr>
<tr>
<td>10020</td>
<td>1.1</td>
<td>0.3</td>
<td>1.4</td>
</tr>
</tbody>
</table>

* Average of four tests.
specimen in Table 6. Histograms of both source parameters are shown for both quench rates (Fig. 21). The emission source volumes from water-quenched samples were slightly larger on average than those from iced-brine quenched samples. There was little significant difference in source lifetimes. The emission activity parameter shown in Fig. 19 was the calculated maximum source volume.

**DISCUSSION**

Variation of specimen heat treatment caused changes in acoustic emission activity during narrow band tests. This, the authors believe, is a consequence of changes in the detailed deformation and fracture mechanisms brought about by microstructural variation. The authors will briefly examine the variation in deformation behaviour and then deduce which of these processes was the origin of detectable signals. Finally, the wide band results will be used to obtain quantitative data about the most energetic emission in these materials.

**Deformation and fracture mechanisms**

General yield is characterized by the widespread propagation of dislocations. The transition in yield-point behaviour from quenched to either slow-cooled or heavily tempered conditions (Figs. 10 and 12) is indicative of changes in the mechanism of motion. In the quenched state, yield is accommodated by movement of the quenched-in dislocation structure, while in tempered/slow-cooled samples, where the dislocation density is initially very low, deformation is accommodated by the nucleation and rapid propagation of fresh dislocations.

In martensitic microstructures the yield stress is expressed as a sum of strengthening components associated with the matrix friction stress $\sigma_0$, the lath packet size $D$, the lath size $d$, and the forest dislocation interactions.$^{11-13}$ Norström$^{13}$

<table>
<thead>
<tr>
<th>Table 6</th>
<th>Wide band acoustic emission measurements</th>
</tr>
</thead>
<tbody>
<tr>
<td>Specimen</td>
<td>Quench method</td>
</tr>
<tr>
<td>Y1</td>
<td>Iced-brine</td>
</tr>
<tr>
<td>Y2</td>
<td>Quench</td>
</tr>
<tr>
<td>Y3</td>
<td></td>
</tr>
<tr>
<td>Y4</td>
<td>Water quench</td>
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<td>Y5</td>
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<table>
<thead>
<tr>
<th>Table 7</th>
<th>Crack length and velocity results: see Fig. 22</th>
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<tr>
<td>Specimen</td>
<td>Quench method</td>
</tr>
<tr>
<td>Y1</td>
<td>Iced-brine</td>
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<tr>
<td>Y2</td>
<td>Quench</td>
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<tr>
<td>Y3</td>
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<tr>
<td>Y4</td>
<td>Water quench</td>
</tr>
<tr>
<td>Y5</td>
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</table>
has assumed that a Hall–Petch relation exists for both lath and lath packet sizes so that the expression for the yield strength of low-carbon martensite is

$$\sigma_y = \sigma_0 + \sigma_1 + k_y D^{-1/2} + k_s d^{-1/2} + \alpha\mu\rho^{1/2}$$

where \(\sigma_0\) is the solution hardening (here due to both carbon and metallic alloying elements), \(k_y\) and \(k_s\) the grain-size coefficients for the lath packet and lath grain sizes, \(\rho\) the dislocation density, \(\alpha\) the dislocation strengthening coefficient, and \(\mu\) the shear modulus. Expressions such as equation (5) for martensitic steels assume independent additive contributions to strength from the components of the microstructure and the existence of a Hall–Petch relation for the grain-size parameters. Both assumptions are difficult to test rigorously but this is of limited relevance here: their importance is that \(k_y\) and \(k_s\) are non-zero, indicative of dislocation interactions both with the lath and lath packet boundaries. Thus, a few dislocations, at least, can propagate a distance comparable to the width of a lath, i.e. several micrometres.

An additional strengthening mechanism in moderately cooled or tempered samples arises from the presence of a dispersion of hard precipitates \(\{\text{M} \text{C}_6, \text{M}_3\text{C}\}\). The magnitude of this strengthening depends upon the size and separation of the precipitates. Gladman et al.\(^{14}\) have shown it to be

$$\sigma \propto \frac{\mu b}{\lambda} \log \left( \frac{\bar{x}}{2b} \right)$$

where \(\bar{x}\) is the mean planar intercept diameter of the precipitates and \(\lambda\) is the surface to surface precipitate spacing given by

$$\lambda = n_s^{-1/2} - \bar{x}$$

where \(n_s\) is the number of precipitates per unit area of slip plane. Assuming a fixed volume fraction of precipitates, the greatest strengthening due to this process occurs for a fine dispersion of small particles, e.g. as found in the air-cooled sample. In the absence of cross slip, the presence of a dispersion of impenetrable precipitates will limit the distance dislocations can freely propagate to \(~\lambda\). The presence of a dispersion of precipitates results in an enhanced work-hardening rate since the dislocation loop left around a bypassed precipitate reduces the effective interparticle spacing. This in part is the origin of the increase in \(n\) value with decrease in cooling rate.

The furnace-cooled material contained fine pearlite in addition to dispersion-strengthening carbides. The

<table>
<thead>
<tr>
<th>Table 8</th>
<th>Comparison of mean crack length and velocity data</th>
</tr>
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<tbody>
<tr>
<td>Material</td>
<td>Quench method</td>
</tr>
<tr>
<td>Pure A533B</td>
<td>Iced-brine quench</td>
</tr>
<tr>
<td>Pure A533B</td>
<td>Water quench</td>
</tr>
<tr>
<td>Mild steel(^{9})</td>
<td>Water quench</td>
</tr>
</tbody>
</table>
dispersion-strengthening contribution to the flow strength should be much less in this case since there is a lower volume fraction of carbides within ferrite grains (most carbon is in the pearlite). Within the pearlite, deformation of both cementite and ferrite will occur. Porter et al.\(^{15}\) have shown that the cementite layers within fine pearlite fracture by a ductile necking process.

**Post-maximum stress**

This was the region where the load was falling since the effect of reduction in specimen cross-section due to necking exceeded the increase in strength due to work hardening. In all specimens it was in this region that subcritical microcracks and void growth and coalescence occurred, ultimately causing failure when the remaining uncracked ligaments were no longer capable of supporting the applied load.

In the most rapidly quenched specimens the true stress reached very high levels, \(\sim 2\ \text{GN m}^{-2}\) (Fig. 11), and cleavage microcracking occurred (Fig. 14b). Fractographic observations indicated that cracks made small orientation changes at martensitic lath boundaries. Other studies\(^{16,17}\) have shown that the major change of crack orientation at lath packet and prior austenite grain boundaries effectively stops a propagating microcrack. The authors thus expect individual microcracks to be approximately a lath packet in size and, on average, to be transverse to the tensile axis. It is likely that some plasticity accompanies the growth of these microcracks which may cause blunting after crack arrest at a boundary. Water quenching produced a slightly lower strength martensite and much less cleavage was observed (Fig. 14c). Here, the subcritical cracking occurred mostly by shear which resulted in a zigzag fracture topography as shear separation occurred on alternate planes. This mechanism of fracture has been observed in other steels of high strength and limited work-hardening capacity.\(^1\) It is thought to occur when the deforming ligament between a pair of large voids (perhaps formed at inclusions) exhausts its work-hardening capacity, resulting in a local intense shear band and the creation and coalescence of a microrvoid sheet. Slower cooling rates resulted in the formation of cup and cone fracture (Fig. 14e). The period of falling load now characterized a time when voids previously nucleated at carbides grew and new voids were nucleated after high elastic strain at carbides. The voids formed at carbides may have been nucleated by decohesion of the carbide/matrix interface or by cracking of the precipitates themselves. As deformation continued the ligaments between voids internally necked down and void coalescence occurred.

Tempering of water-quenched specimens, even for only 6 min, resulted in the disappearance of alternating shear and the initiation of cup and cone fracture (Fig. 15). Extending the tempering time increased the size of the ductile dimples on the fracture surface as the number of carbides present decreased although their size increased. There was fractographic evidence that carbide fracture did occur but was mostly confined to regions which had experienced intense plastic strain.

**Origin of detectable acoustic emission**

It is likely that all mechanisms of deformation and fracture dissipate some energy in the form of elastic waves. However, many studies in the past have clearly shown that the elastic waves from some mechanisms are more easily detected than those from others.\(^2\) This study attempts to determine the factors controlling the generation of detectable acoustic emission in a steel within the A533B composition range. If one ignores the effects of ultrasonic attenuation there are
two practical factors which are likely to have a major influence on the detectability of an acoustic emission. One is the sensitivity of the detection system to the surface displacements caused by arrival of elastic waves from an emission source. Narrow band systems with their better signal-to-noise characteristics have a much greater sensitivity compared with wide band techniques. The other is the frequency response of the detection system compared with the spectrum of the source.

**Sensitivity limitations**

All detection systems have some kind of detection threshold and only acoustic emission signals whose amplitudes exceed this are recorded. The nature of the threshold depends on the type of acoustic emission instrumentation in use. In systems where ring-down counts are made, the threshold is a voltage level which the emission signals must exceed and may correspond to a minimum detectable surface displacement or velocity of the specimen surface depending on the mode of operation of the transducer. The wide band system used in this study had a voltage threshold in the transient recorder (set to be just above the noise level) which corresponded to a minimum detectable surface displacement of $10^{-12}$ m. The narrow band system, however, had no such clearly defined threshold. Here, the transducer output was monitored continuously and the effective minimum detectable signal was that producing the smallest power fluctuation which could be distinguished from noise. This corresponded to a frequency-band limited power rather than a displacement, so that the minimum detectable displacement was frequency dependent. However, in order to simplify the following discussion the authors assume it to be constant at $\sim 10^{-14}$ m. We can then proceed to compare this value with the displacements from dislocation motion and crack-growth sources.

**Dislocation sources**

The authors consider the expansion of a glissile dislocation loop of Burgers vector $b$ from an initial radius of zero to $a$. It is assumed the loop is inclined at 45° to the tensile axis of the specimen and that the final loop radius is much less than the specimen surface. The time dependence of the surface displacement at the epicentre of a half-space produced by such a source has already been evaluated. 1 The early part of the wave form is dominated by a pulse due to the arrival of the longitudinal wave; the area under this pulse is
given by

$$S_d = \frac{c_1^2 b A}{2 \pi h c_1^3}$$

(8)

where $h$ is the depth of the loop below the transducer and $A$ is the final loop area.

If the loop grows at constant radial velocity $v$ so that the duration over which it grows

$$\tau = a/v$$

(9)

then $\tau$ will correspond to the width of the longitudinal wave pulse. If the height of the pulse is $x$, then for a loop area that grows parabolically in time to its final value of $\pi a^2$, the surface displacement first arrival commences with a ramp of slope $x/\tau$. Thus

$$x = \frac{h v b c_1^2}{2 c_1}$$

(10)

If $x_{\text{min}}$ is the displacement detection threshold, the smallest detectable dislocation loop must have grown to a radius

$$a_{\text{min}} = \frac{h c_1^2 x_{\text{min}}}{b v c_1^2}$$

(11)

For the detection system used here $x_{\text{min}} \approx 10^{-4} \text{ m}$. For steel, $c_1 = 5960 \text{ m s}^{-1}$, $c_2 = 3200 \text{ m s}^{-1}$, and an average source depth $h = 0.03 \text{ m}$. Using a Burgers vector of $\sim 3 \times 10^{-10} \text{ m}$ and assuming the dislocation velocity was $200 \text{ m s}^{-1}$, $a_{\text{min}} \sim 100 \text{ m}$. Thus, if a single dislocation loop were to expand to a radius of $\sim 100 \text{ m}$ at a radial velocity of $200 \text{ m s}^{-1}$ it would be just detectable. Loops of smaller radius could be detected according to the model if they either grew at greater velocity or if $n$ loops situated close to each other expanded together over the same time interval so that their combined radii exceeded 100 m. In the latter case

$$a_{\text{min}} = \frac{h c_1^2 x_{\text{min}}}{b n v c_1^2}$$

(12)

**Crack-growth sources**

It is assumed that the source is the creation of a small horizontal microcrack of cross-sectional area $A$, crack face separation $2\delta$, and volume $V$, under mode I loading. Then at a distance $h$, much greater than the crack dimension, the strain field approximates to that of an edge dislocation loop of area $A$ and effective Burgers vector $2\delta$. The area under the longitudinal wave pulse is given by

$$S_e = \frac{V}{2 \pi c_1 h}$$

(13)

Assuming a parabolic increase in crack area with time, the peak amplitude of the longitudinal component is

$$x = \frac{V}{\pi c_1 \tau h}$$

(14)

This enables the minimum detectable crack volume to be estimated, provided $\tau$ is known. However, it is more usual to work in terms of crack length. Equation (14) can be expressed in terms of length if it can be assumed that the crack behaves in a linear elastic manner during its growth. In this case, the crack will have an ellipsoidal shape of volume

$$V = \frac{4 \pi a^2 \delta}{3}$$

(15)

where $a$ is the radius of an assumed circular crack of midplane area $\pi a^2$. Under elastic loading the crack face displacement at the centre of the crack $\delta$ is a function of crack radius and applied stress

$$\delta = \frac{2(1 - \nu^2) \sigma a}{E}$$

(16)

where $E$ is Young's modulus and $\nu$ is Poisson's ratio. Substituting for $V$ and $\delta$, equation (14) becomes

$$x = \frac{8 a^3 (1 - \nu^2) \sigma}{3 c_1 \tau h E}$$

(17)

Assuming a constant radial crack velocity $v$ so that $a = vt$ and rearranging equation (17) to give

$$a_{\text{min}} = \frac{3 c_1 E h x_{\text{min}}}{8 (1 - \nu^2) \sigma v}$$

(18)

From equation (18) it is seen that the smallest crack advances are most easily detected when they are close to the
transducer, of short duration (high velocity), and occur in a high-strength matrix.

Using \( E = 2 \times 10^{11} \text{ N m}^{-2}, \quad v = 500 \text{ m s}^{-1}, \quad v = 0.3, \quad \sigma = 10^9 \text{ N m}^{-2} \) and \( h, c_1, \) and \( x_{\text{min}} \) as before, the authors find that the smallest detectable crack with the narrow band system has a diameter \( 2a_{\text{min}} = 10^{-6} \text{ m}. \) \( a_{\text{min}} \) has a different dependence on \( x_{\text{min}} \) for dislocation loop and crack-growth sources (equations (12) and (18)). The relative sensitivity

\[
\frac{a_{\text{min}} \text{ (dislocation)}}{a_{\text{min}} \text{ (crack)}} \propto x_{\text{min}}^{1/2}
\]

Thus, for this model, it would be expected that as a detection system was made more sensitive, it would preferentially detect small dislocation loop sources. A less sensitive detection system such as the broad band system \( (x_{\text{min}} = 10^{-12} \text{ m}) \) should more readily detect crack-growth sources than dislocation loop sources.

**Band width limitations**

All acoustic emission detection systems have a limited frequency response. The most serious consequence is that the wave form is distorted so that dynamic information about the source is lost. Thus, the wide band system was developed to have the widest possible band width to enable measurement of dynamic source properties. There is, however, a second effect of a limited frequency response: it leads to a reduction in effective sensitivity of a recording system, dependent upon the band width of the detection system relative to the spectrum of the source. The narrow band system was band-limited to the frequency range 0.1-1.0 MHz (and many other systems in current use are even worse). It is therefore important to examine the effect of this upon sensitivity.

For simplicity, we assume that the emission pulse is Gaussian of width \( \tau \) and height \( x \). Its area is \( k\tau x \) where \( k \) is a constant. First examine the effect of low-pass filtering. Suppose the pulse is filtered by a filter whose impulse response is the Gaussian pulse of width \( \tau_1 \) and unit area, i.e.

\[
(1/\tau_1 \sqrt{\pi}) \exp(-t^2/\tau_1^2)
\]

The effect of filtering, expressed mathematically by convolution in the time domain, is to generate a further Gaussian pulse as the output, of width \( \sqrt{\tau^2 + \tau_1^2} \) and height \( x' \).

The areas of the Gaussian pulses multiply so that the area of the input and output pulses are equal, i.e.

\[
k\tau_1 x = k\tau x' \sqrt{\tau^2 + \tau_1^2}
\]

Thus, by rearranging equation (20) and noting that \( \tau \approx \tau_1 \)

\[
x' \approx \frac{x\tau_1}{\tau} \quad \text{ (21)}
\]

The detected displacement is reduced by a factor equal to the ratio of the filter frequency to the bandwidth of the incoming signal.

Now consider the same incoming pulse filtered by a high-pass filter which can be represented as the complement of a second Gaussian of unit area and width \( \tau_2 \), where \( \tau_2 \) corresponds to the corner frequency of the filter, i.e. the filter function is

\[
\delta(t) = \frac{1}{\tau_2 \sqrt{\pi}} \exp(-t^2/\tau_2^2)
\]

When convolved with the input pulse this gives the difference between two Gaussian functions, the first equal to the input and the second of equal area, but of width \( \sqrt{\tau^2 + \tau_2^2} \) and height \( x', \) i.e.

\[
k\tau x' = k\tau x' \sqrt{\tau^2 + \tau_2^2}
\]

The height of the output pulse is therefore

\[
x = x' \left[1 - \frac{\tau}{\sqrt{\tau^2 + \tau_2^2}}\right] \quad \text{ (23)}
\]

For \( \tau_2 \ll \tau \)

\[
x = x' \frac{x_0^2}{2\tau^2} \quad \text{ (24)}
\]

Due to the high-pass filter the height of the pulse is reduced by a factor \( 0.5 (\tau_2/\tau)^2 \), proportional to the square of the ratio of the band width of the signal to that of the filter.

A pulse whose width was 100 ns (a typical value for an emission from a crack) subjected to a 1 MHz low-pass filter \( (\tau_1 = 0.35 \mu s) \) will have its amplitude attenuated by a factor of 0.27 (from equation (21)). Furthermore, the degree of this attenuation will vary with the band width of each emission. Similarly, a broad pulse of 10 \( \mu s \) width will suffer significant amplitude attenuation when subjected to a 100 kHz high-pass filter \( (\tau_2 = 3.5 \mu s) \). From equation (24) the amplitude is reduced by a factor 0.06.

The narrow band detection system thus tends to underestimate, and may even fail to detect, signals whose band width extends beyond the pass band of the detection system.

Further complications also exist due to reflections from boundaries of the specimen. These are difficult to analyse and are ignored here. Their main effect will be to increase, somewhat, the probability of detecting a source. In the qualitative discussion to follow the authors assume that effects due to reflections are approximately cancelled by those due to limited band width.

In summary, while narrow band techniques have high sensitivity and are thus able to detect weak signals, they distort the acoustic wave form so severely that little useful information can be deduced about the source. Conversely, wide band measurements record only energetic events but with sufficient fidelity to facilitate source characterization.

**Detectable acoustic emission sources in A533B steel**

The above considerations have indicated limitations of acoustic emission techniques. Because of these, not all deformation and fracture processes that occur in a metal are likely to be detectable. In the material studied here, a wide range of deformation and fracture processes were induced and the authors will now attempt to determine which of these were responsible for the observed emission. This is greatly simplified by the exclusion of the MnS stringers from this material since these may be a source of acoustic emission that could mask the emission from other processes.

**Premaximum stress**

The four most rapid quench rates resulted in no significant emission during this period of deformation. The few isolated and random emissions that were observed could have been the decohesion–cracking of spheroidal inclusions. It appears that deformation processes in material in these conditions generate no detectable emission. This is probably because the high dislocation density and the small width of martensite laths result in very short distances for dislocation propagation, well below the detection limit of \( \sim 100 \mu \text{ m} \) for a single dislocation. Air and furnace cooling gave increased levels of continuous emission around yield. The occurrence of these correlates with an increase in the distance of
dislocation propagation, due to the disappearance of the lath structure and a reduction in initial dislocation density.

The tempering of water-quenched specimens had a greater effect. Lightly tempered martensites generated little or no emission which was consistent with dislocation pinning by other dislocations, carbides, and lath boundaries. Prolonged tempering, however, resulted in a strong increase in emission as the quenched-in dislocation structure and the lath boundaries were annealed-out, and the size and carbide spacing increased. The calculations above estimate that a single detectable dislocation loop would need to expand to a radius of at least ~100 μm if its velocity was ~200 m s⁻¹. In the quenched condition the lath width was ≤10 μm, so even at a velocity of 200 m s⁻¹ a very large number of dislocations would have to move cooperatively in order to generate a detectable signal. Short tempering treatments, by introducing small impenetrable carbides while not greatly changing dislocation density or lath structure, would tend to reduce still further the effective distance of propagation between pinning events. In the specimens tempered for long times many of the barriers to dislocation motion are removed or their effect reduced, resulting in the detection of many very small amplitude emissions for a short interval of strain around yield. Forest interactions during work hardening presumably limit dislocation propagation distances during post-yield straining and lead to a progressive disappearance of the emission.

Post-maximum stress

Varying the cooling rate of quenched specimens markedly affected the emission activity during post-maximum stress deformation (Fig. 16). The two fastest quenched specimens generated very large amplitude emissions as final fracture approached. This period of deformation was characterized by the formation of subcritical cleavage cracks. If the cracks were a lath packet in length (~30 μm), then for these steels emissions of amplitude about 30 times the threshold would be expected. It thus appears that the most likely source of emission was the formation of cleavage microcracks and that the strain dependence of the detected emission was an indication of the rate of cleavage microcrack formation.

When the cooling rate was reduced slightly (by quenching in water) the emission activity was both reduced in energy and delayed, occurring closer to final failure (Fig. 16c). Here the fracture mode was alternating shear with only a few isolated areas of cleavage. Both fracture processes could generate detectable emission. The areas of cracking from each were sufficient to generate signals whose amplitude would greatly exceed the threshold value. The narrow band results provide insufficient evidence to determine whether one or other or both were the sources of emission, although the reduction in emission may be an indication that in these materials the alternating shear processes generate weaker signals than cleavage.

Neither the slow cooled nor any tempered specimen generated any emission other than at final failure. These specimens fractured essentially by ductile void coalescence. However, the wide range of heat treatments resulted in a wide range of void densities and sizes. These results are consistent with those of earlier studies of a 4Ni-Cr steel and indicate that for a wide range of microstructures void growth and coalescence (which involve no long-distance propagation of individual dislocations or sudden fracture over several micrometres) generate no detectable acoustic emission under uniaxial loading.

Carbide cracking occurred in the samples containing large carbides (Fig. 8). These were M₂₃C₆ and M₆C in composition (Fig. 9) and ≤2 μm in size. Signals from the fracture of these particles are just on the limit of detectability if the fracture occurs in a brittle manner. The absence of acoustic emission at the large plastic strains when fracture is expected to have occurred indicates that cracking of these composition carbides is not a major acoustic emission source.

Acoustic emission characterization of microcrack growth

The acoustic emission results discussed above have shown that only a few of the wide range of deformation and fracture events are detectable, even when using sensitive recording techniques. Because of the severe distortion experienced by the signals during the sensitive narrow band tests, these can provide only a qualitative description of emission source activity.

However, selecting only those specimens that emitted large-amplitude emission signals, i.e. the iced-brine and water-quenched specimens, so that the less sensitive broad band system could be used, it is possible to measure emission wave forms sufficiently free from distortion to obtain a single parameter description of the source in terms of a time-varying crack volume (Table 6). The results showed distributions of source volume and lifetime for both heat treatments (Fig. 21). It is noted that typical lifetimes are ~100 ns, so that the wave forms have significant frequency content outside the band width of all except wide band detection systems. Thus, with narrow band systems there are likely to be the errors in amplitude estimation outlined above.

It is more usual to deal with crack length rather than volume for crack-growth events. The authors calculate crack length for each recorded wave form making the same assumptions, principally that the crack is horizontal, circular, and under mode I elastic loading, and using the same equations (13)-(18) as above. A second useful parameter, the crack velocity, is calculated by dividing the crack diameter by its lifetime, assuming uniform radial velocity of the crack front. The results are plotted as histograms in Fig. 22 and the mean values (with standard deviations also given as a guide to the widths of the distributions) tabulated in Table 7 which shows consistency between the results of individual specimens, but with somewhat greater deviations for the specimens with the poorest statistics.

By comparing the histograms for the two different cooling rates and the mean crack lengths and velocities (Table 8) it can be seen that while the slightly slower cooling rate gives markedly fewer emissions, it gives virtually indistinguishable distributions of crack length, lifetime, and velocity. It was concluded above that the source of emission during post-maximum stress deformation in iced-brine quenched material was cleavage fracture. In the case of water-quenched material there was some doubt about whether the source was again cleavage or alternating shear fracture. The quantitative emission data show no difference within statistical error between the two materials, suggesting that the emitting crack-growth process is the same in both materials. Furthermore, the estimated total detected crack areas (calculated by summing the individual microcrack areas) is ~4 and ~1% of the gauge section for iced-brine and water-quenched specimens respectively. The low detected area in the latter case would be consistent with fractographic evidence of a lower incidence of cleavage microcracking. This is an example of the extra power of
CONCLUSIONS

1. The origin of yield-region acoustic emission from Mn–Mo–Ni A533B steel containing a low concentration of sulphur appears to be the motion of groups of dislocations whose combined slip distance exceeds about 100 μm. Such events are restricted to slowly cooled or heavily tempered microstructures as these have the most widely spaced dislocation barriers. Carbide cracking does not generate detectable signals in this alloy system in the heat-treated conditions investigated.

2. Acoustic emission before fracture appears only if cleavage or possibly alternating shear microfracture occurs. The wide range of ductile dimple fracture modes produced in this steel generate no detectable emission over the frequency range 0.1–1.0 MHz.

3. Quantitative broad band recording techniques coupled with time domain inversion of measured wave forms enabled the emission source during fracture to be characterized. This source corresponds typically to a cleavage microcrack propagating ~50 μm at an average speed of ~450 m s⁻¹.

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