Inclusion deformation and toughness anisotropy in hot-rolled steels

The basic factors controlling the deformation behaviour of non-metallic inclusions during the hot working of steels are discussed with particular reference to silicate inclusions, alumina, and the three different types of MnS inclusion. The factors to be considered include: the initial morphology of the inclusion; the inclusion composition; the strength (hardness) of the matrix and inclusion phases at hot-working temperatures; the inclusion size; the strength of the matrix/inclusion interface; the degree of deformation; associated void formation. The effects of inclusion deformation and reorientation on the short-transverse toughness of hot-rolled steels are reported, with particular reference to steels containing Types I, II, and III MnS. The effects of inclusion volume fraction, aspect ratio, distribution, size, and interinclusion spacing are examined and the mechanism of short-transverse fibrous fracture is discussed in order to identify the important factors controlling toughness. MT/198D

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The detrimental effects of non-metallic inclusions on the ductility and toughness of both cast and wrought steels are now well established. The important features of the inclusion dispersion which can affect toughness are the size and distribution of the inclusions, their total volume fraction, and the shape that they acquire as a result of the prior deformation of the matrix. In the first part of this paper, the basic factors controlling the deformation of non-metallic inclusions in steel during a working process are considered with particular regard to recent work in which the deformation of inclusions is related to relative yield behaviour (hardness) of inclusions and matrix. In the second part of the paper, the influence on toughness of these various factors is examined with particular reference to experimental work into the effects of the various types of manganese sulphide inclusions on the short-transverse toughness of hot-rolled steel. The mechanism of fibrous fracture under these conditions is also discussed in order to identify the main factors which control toughness.

Factors controlling inclusion deformation

During the hot working of steel containing non-metallic inclusions, stresses are exerted on the inclusions by the steel matrix which tend to make them deform in a manner similar to that of the matrix. Thus inclusions that are initially approximately spherical, Types I and III manganese sulphide and silicate inclusions for instance, will deform during rolling to give an ellipsoidal geometry. Type II manganese sulphide and alumina clusters present an entirely different starting condition, being present in the cast steel as interconnecting colonies. During rolling, these whole colonies rotate into the rolling plane and subsequently elongate, giving rise, as will be discussed below, to a very much reduced toughness in the short-transverse direction.

The extent to which an inclusion deforms depends primarily on the ratio of the flow stress of the inclusion to that of the matrix, a ratio which is itself a function of temperature and composition of the two phases. Other factors such as interfacial energy and mechanical constraint do exert an influence, particularly at large reductions, if the inclusion becomes very elongated. These various factors will now be considered in more detail.

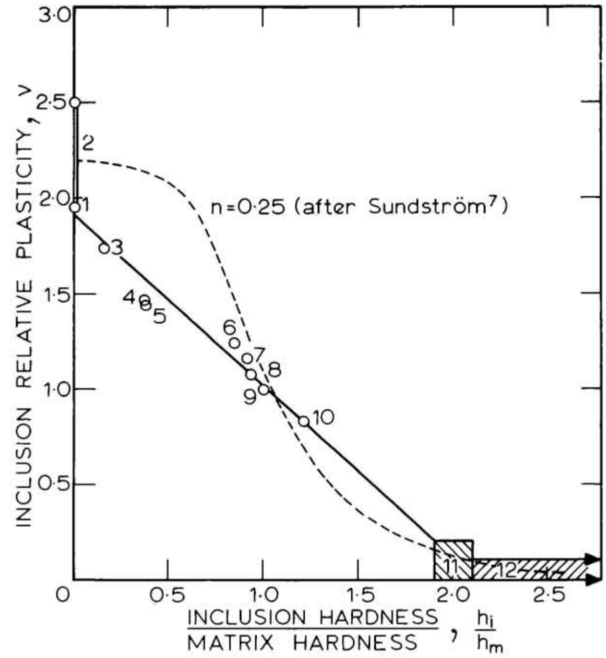
RATIO OF FLOW STRESSES

Using a novel high-temperature microhardness tester[1] it has been possible to determine the indentation hardness of non-metallic inclusions *in situ* together with that of the surrounding matrix up to a temperature of 1000°C. Indentation hardness cannot of course describe all the variables associated with the yielding process, but it would clearly be impracticable to measure any parameter other than hardness on inclusions *in situ*. Moreover, it has been shown[2] that for most isotropic materials, hardness is directly related to the flow stress at 8% strain (the exact value of strain depending on the indenter geometry). Consequently, hardness gives a reliable correlation with deformation behaviour at comparatively low strain or under conditions of negligible strain hardening. By measuring the inclusion and matrix hardness at different temperatures for a number of real and artificial inclusion systems and determining the relative plasticities (ν= the ratio of inclusion true strain to overall matrix true strain), data have been obtained[3,4] (Fig.1 and Table 1) to yield the following relationship:

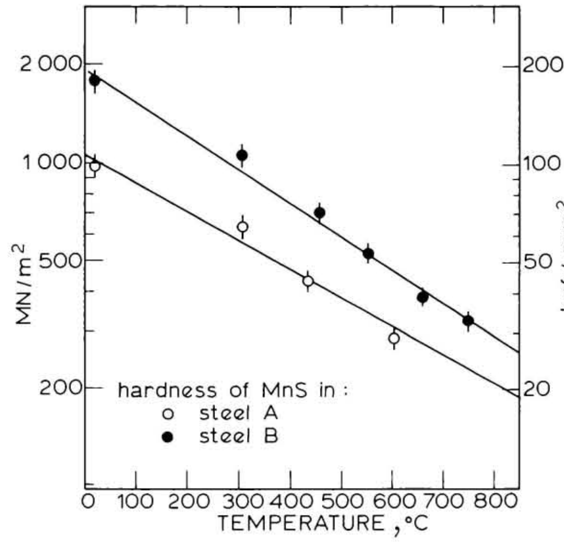
$$\nu \simeq 2 - h_i/h_m \ (0 \leq \nu \leq 2)$$

where h_i is the inclusion hardness and h_m is the matrix hardness. It should be pointed out that relative plasticity invariably decreases with increasing overall reduction when other factors, to be discussed below, become significant. The relative plasticity referred to here is obtained by extrapolation to zero deformation of the values of ν obtained at varying reductions.

It can be seen that an inclusion will undergo little or no deformation if it is more than twice as hard as the matrix and will extend little more than twice as much as the corresponding matrix strain if it is fluid. The situation of a rigid spherical inclusion within a matrix subjected to a uniaxial stress σ has been analysed[5] for the elastic condition and the maximum stress acting on the inclusion shown to be 2σ. Because of strain hardening, however, it is possible for inclusions that are more than twice as hard as the matrix to undergo a little deformation when the matrix



1 Relationship between relative plasticity and the ratio of inclusion to matrix hardness



2 Hardness of Type I MnS inclusions in relation to temperature

strain is large. With regard to fluid inclusions, measurements of water-filled voids in plasticine and of siliceous inclusions deformed in the fluid state normally give values of the order of 2 for relative plasticity. The joining of fluid silicate inclusions on very close approach[6] can, however, give rise to misleading results.

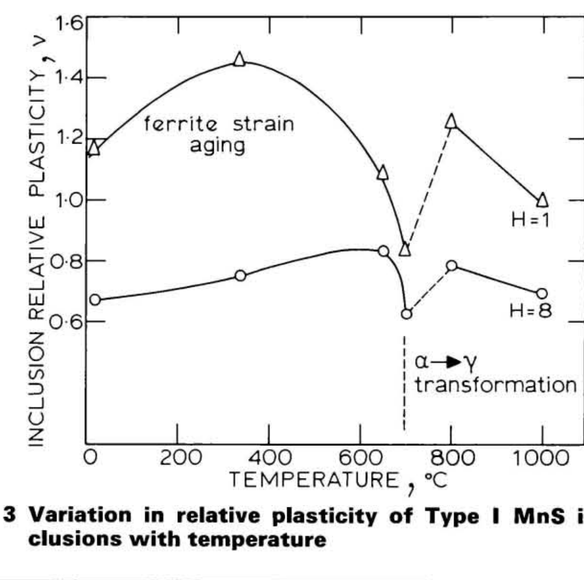
Sundström[7] has reviewed and extended the visco-elastic theoretical analyses in relation to plastic inclusions and matrix, giving graphical relationships between relative strain in the two phases and the ratio of flow stresses for various values of the work-hardening exponent (n), apparently assumed to be the same for both materials. While the experimentally determined values give a straight-line relationship, the Sundström model predicts a curve, the form of which depends on the value of n chosen. In a realistic situation there will inevitably be some deviation from the assumption about work-hardening exponents. A number of workers have carried out experiments using artificial inclusion systems to obtain similar data. Zeisloft and Hosford[8] using a Wood's metal–tin–lead system have obtained graphical relationships similar to those of Sundström, with a work-hardening exponent of 0·25, except for some divergence of very soft inclusions giving a higher maximum relative plasticity nearer 3. Warrick and Van Vlack,[9] extruding sintered metals containing ionic inclusions of the NaCl-type structure, obtained data that would indicate a maximum relative plasticity of about 2.

Table 1 Relationship between inclusion behaviour and relative hardness

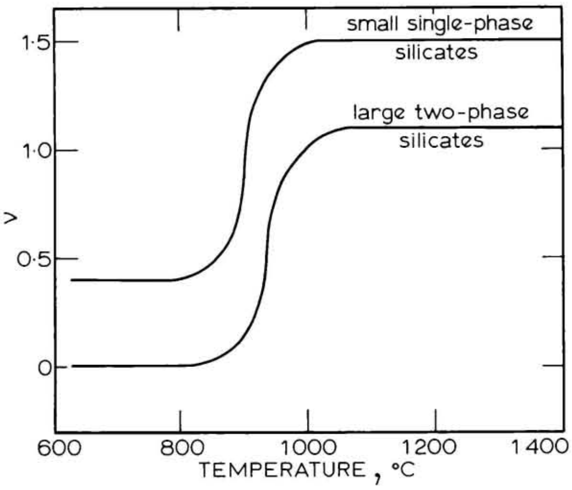
	Relative plasticity	Relative hardness, H_i/H_m	Type of inclusion system
1	1·95	0	Water-filled holes in plasticine
2	1·9–2·0	0	Silicate inclusions > 1 000 °C
3	1·74	0·16	Copper rods in steel matrix 900 °C
4	1·46	0·38	Copper rods in steel matrix 700 °C
5	1·45	0·38	MnS inclusions 340 °C
6	1·25	0·85	MnS inclusions 800 °C
7	1·16	0·91	MnS inclusions 20 °C
8	1·08	0·93	MnS inclusions 650 °C
9	1·00	1·00	Steel rods in steel matrix 900 °C
10	0·83	1·21	MnS inclusions 700 °C
11	0–0·2	1·9–2·1	Silicate inclusions at 800 °C
12	0–0·1	> 2·0	Silicate inclusions <800 °C Tungsten rods in steel 700 °C

TEMPERATURE

Variation of temperature affects the hardness of both the inclusion and matrix. Also, temperature-induced structural changes of the matrix, such as the ferrite–austenite transformation in steels, can give rise to rapid changes in the relative plasticity of inclusions. At a lower temperature the dynamic strain-aging peak in the matrix hardness will also increase the relative plasticity. Figure 2 shows the hardness variation with temperature of certain Type I MnS inclusions together with that of the ferritic matrix in which they were situated, and Fig.3 shows the corresponding changes in relative plasticity. A peak in relative plasticity occurs at the lowest temperature where the matrix is wholly austenitic, and the decrease in relative plasticity as the temperature increases to 1 200°C is now well established.[6,10] With regard to silicate inclusions the general shape of the curves of relative plasticity v. temperature are shown in Fig.4.[6] The indentation hardness of manganese



3 Variation in relative plasticity of Type I MnS inclusions with temperature



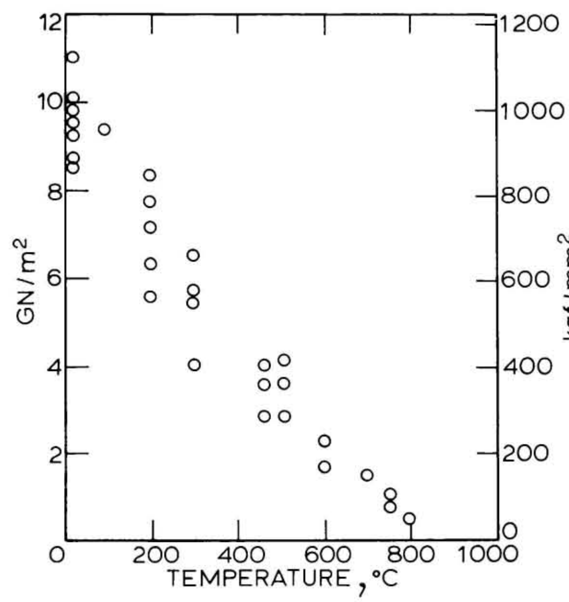
4 Relative plasticity of silicate inclusions

aluminosilicate inclusions has been determined over a range of temperature[3] (Fig.5) and, as expected, their hardness is very much greater than that of the steel matrix at low temperatures but falls rapidly to zero at the temperature corresponding to the transition from rigid to fluid behaviour.

CHEMICAL COMPOSITION

In the case of single-phase manganese sulphides, the high-temperature hardness is a function of solute content. The presence of oxygen and also of calcium has been shown[11] to increase the high-temperature hardness of bulk synthetic manganese sulphide, and increased hardness with higher oxygen content has been noted for *in situ* MnS inclusions.[3] This effect of oxygen is of particular significance in relation to the three types of MnS where the intrinsic hardness falls from Types I to II to III with a corresponding increase in relative plasticity.

The effect of increasing oxygen content on the relative plasticity of Type I MnS is to move the curve in Fig.3 down to lower levels of ν. However, at high oxygen contents,



5 Hardness of silicate inclusions

where Type I sulphides give way to duplex inclusions containing MnO, the high-temperature behaviour becomes more consistent with that of oxides, where there is a sharp increase in ν at about 1 100°C consistent with the melting point of the MnO/MnS eutectic.

Chromium, which replaces manganese in the sulphide, has been said to give an increase in hardness,[12] and it is also known that the addition of zirconium, titanium, cerium, or other rare-earth elements results in a sulphide of considerably reduced relative plasticity.[13] In these cases it is also likely that the oxygen content is significant in affecting the intrinsic hardness. In this connection, it should be recognized that there is a danger in adding an element to combine with sulphur which is also a strong deoxidizer. If all the sulphide is not converted to, say, CeS, because the additions are inadequate in relation to the oxygen content, the MnS remaining may be of the Type II form. The final inclusion assembly may then be little better or even more harmful than would otherwise have been the case, in that directionality is rapidly produced on rolling.

With regard to silicate inclusions, it has been shown that the temperature of sharp transition between rigid and fluid behaviour is controlled by the composition of the silicate glass which is determined by the deoxidation practice.[14]

INITIAL SIZE OF THE INCLUSION

The bulk of practical observation suggests that deformation decreases with inclusion size. If the inclusion is considered initially to be a cube of side d, deformed to $Hd \times d/H \times d$, it can be shown[10] that the ratio of interface energy to deformation energy is

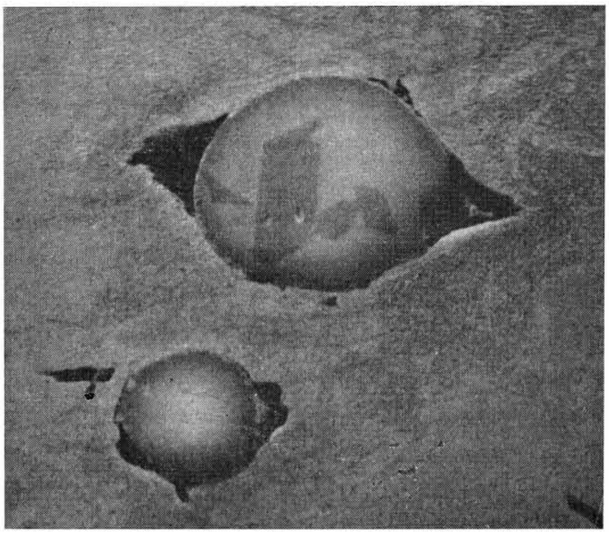
$$\frac{E_{\mathrm{int}}}{E_{\mathrm{def}}} = \frac{2\gamma(H-2+1/H)}{Y d \log_e H}$$

where γ is the interfacial energy and Y is the compressive stress on the deformable inclusion, assumed to be equal to the yield stress of the matrix. This implies that a reduction in inclusion size will reduce the energy available for deformation. However, the practical observations are complicated by the fact that in any given population there is likely to be a variation in composition with size. Thus, in the case of manganese sulphides, the later formed smaller particles, usually at primary dendrite boundaries, will be lower in oxygen and may even move from a Type I to a Type II morphology.

STRENGTH OF THE MATRIX/INCLUSION INTERFACE

The strength of the inclusion/matrix interface is not generally known, but it is probably low judging by the ease with which voids nucleate at the interface during rolling (Fig.6) and in fracture processes. Also, certain types of inclusions, including MnS, contract more than the matrix on cooling and may be always separated by voids.[15] In relation to inclusion deformation the interfacial strength is probably not high enough to modify greatly normal frictional forces, nor to prevent void formation during the rolling process under conditions of a low-value relative plasticity.

Where the interphase bonding strength is relatively high, as for intermetallic compounds in aluminium, present work by Ashok[16] indicates that there is a significant effect on second-phase deformation, with differing effects from non-metallic inclusions in steel. This is an area where the Sundström mathematical model does not fully describe the system and where the Zeisloft and



6 Void formation at inclusions during rolling ×800

Hosford models[8] could be expected to deviate in behaviour from, say, sulphides in steel.

DEGREE OF DEFORMATION

It is now clearly established that the relative deformation of sulphide inclusions decreases as the overall deformation of the steel increases. As the inclusions become elongated the rate of elongation diminishes and the early stages of rolling produce the most marked changes. This has important consequences in the practical hot rolling of steels containing inclusions such as sulphides whose relative plasticity increases with decreasing temperature. During the initial rolling passes, temperatures are high and inclusion deformation is restricted due to their low relative plasticity. During the final stages of hot rolling, the inherent relative plasticity of the inclusions is high and very large elongations might be expected. However, by this stage the inclusions have already undergone some deformation and this markedly reduces their deformability. Fortunately, therefore, the inclusion aspect ratios developed during the final states of hot rolling are much less than would be predicted on the basis of the 'initial' relative plasticity value. There are three possible reasons for the decrease in relative plasticity with increasing deformation.

Relative strain-hardening rate

The strain-hardening rate of the inclusions could be higher than that of the matrix at the temperature involved. In the as-cast form the sulphides may be single crystals and thus their behaviour during rolling in terms of recrystallization and the generation of polycrystalline assemblies could be non-uniform as influenced by their original orientations.[17] Bonizweski and Baker[18] have identified dislocation structures within sulphide inclusions but Moore,[19] measuring the hardness of sulphides before and after deformation, did not find any increase in hardness, although it is difficult to measure hardness with accuracy on thin sulphides *in situ* after rolling without interference from the matrix. There is not yet enough evidence to dogmatize, but the indications are that manganese sulphide does not work-harden significantly under the normal hot-rolling conditions.

Matrix constraint

Where inclusions have initial relative plasticity values less than unity, a mechanism has been proposed[10] whereby the matrix, in trying to flow over the inclusion, becomes increasingly constrained with increased inclusion deformation by friction at the interface. This impedes deformation and flow of the steel around the inclusion, which in turn produces a corresponding reduction in inclusion deformation. Under such conditions, as deformation proceeds the matrix will become increasingly constrained along the length of the interface and more free to flow at the limits, giving the characteristic ogee-shaped end to inclusions which are observed.

Local work hardening of the matrix

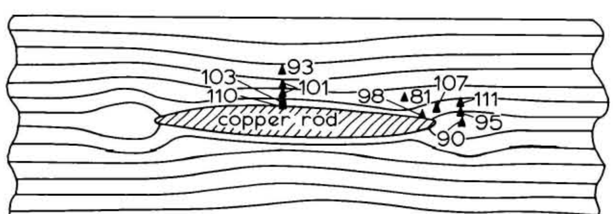
Where the initial value of relative plasticity is above unity, there are also conditions of reduced relative plasticity with increasing H. There are two possible explanations here. Either a sliding of material can occur across the interface with corresponding frictional forces being exerted on the matrix, or, if the interface remains intact, a narrow region of the matrix adjacent to the inclusion can be constrained to deform to the same extent as the inclusion itself. In either event the forces across the interface will tend to impede the inclusion deformation, resulting in a decrease in relative plasticity with increasing reduction, and cause a local matrix strain in excess of the overall strain of the material. At temperatures where this rate of work hardening exceeds the rate of annealing, this region of increased local strain will become harder than the bulk material. Thus, a soft inclusion becomes surrounded by a hardened shell that tends to resist any further deformation. Such an effect has been clearly verified by models and microhardness measurements.[3] Figure 7 shows the flow around copper 'inclusions' in ferrite rolled at 650°C where there is no hardening of the copper but progressive work hardening of the ferrite. It is significant, also, that for sulphides in steel the greatest decrease in relative plasticity is seen to occur below about 500°C, where initial values of ν are high but where rapid work hardening of the matrix occurs.

Factors controlling the effect of inclusions on toughness

Inclusion deformation is the major cause of toughness anisotropy in hot-rolled steel. However, many other factors such as inclusion volume fraction, initial inclusion morphology, and inclusion size also have an important influence, and in the following section each of these factors is examined in detail.

INCLUSION VOLUME FRACTION

In nearly all systems containing a dispersion of second-phase particles it is found that the ductility is strongly influenced by the volume fraction of second-phase particles. This is especially true of non-metallic inclusions in steel



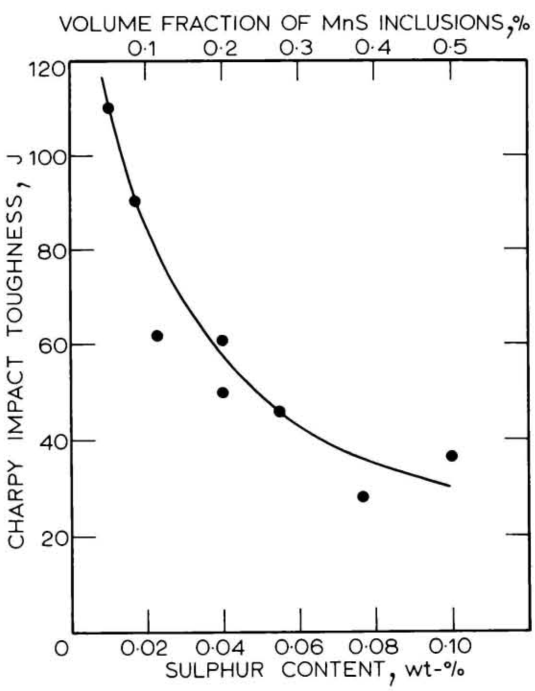
7 Copper 'inclusions' in ferrite showing local work hardening of matrix

and the reciprocal type of relationship between impact toughness and inclusion volume fraction shown in Fig.8 has been demonstrated for many inclusion systems.[20] Likewise, very similar behaviour has been observed for the effect of inclusion volume fraction on tensile ductility[21-23] and plane-strain fracture toughness.[24]

Where inclusions dominate the fracture behaviour, the mechanism of fracture is almost invariably microvoid formation, due to decohesion of the inclusion/matrix interface, followed by void coalescence. Consequently, it might be expected that the strength of the particle–matrix bond would have a prominent effect on the fracture behaviour. In the case of carbide particles in steel[23] which are strongly bonded to the matrix, this is observed. With inclusions in steel, however, most of the available evidence indicates that there is no significant effect of inclusion composition and it is observed that similar dispersions of sulphide, silicate, or alumina inclusions have similar effects on toughness at the same volume fraction.[20] This suggests that all inclusions are only very weakly bonded to the matrix, and as discussed above, this is borne out by the fact that interface decohesion occurs readily during rolling and is observed at very low strains during fracture testing.

Recent evidence by Easterling[25] has suggested that the composition of the steel may affect the strength of the inclusion interface. Thus, under certain conditions, fracture of the inclusion rather than interface decohesion has been observed during straining. However, it must be noted that these experiments were carried out on sintered powder compacts and hence interface reactions have undoubtedly occurred between the inclusion and the matrix, and consequently this may not reflect the behaviour of normal inclusions in bulk steel. Nevertheless, this is an important area where further work is required.

Despite the similarity between the effects of the different types of inclusion, there is no unique relationship between toughness and inclusion volume fraction owing to the powerful influence of the other inclusion variables on fracture behaviour. Thus, although a low inclusion volume fraction is a necessary requisite for good toughness, it is by no means sufficient.



8 Effect of inclusion volume fraction on transverse impact toughness

INCLUSION MORPHOLOGY

Deformation of non-metallic inclusions produces a marked anisotropy in the fracture behaviour of steel. Thus, although the toughness is not reduced in the longitudinal direction, in the transverse and especially the short-transverse directions (when the elongated inclusions are orientated normal to the direction of applied stress) very low toughness properties can be obtained.

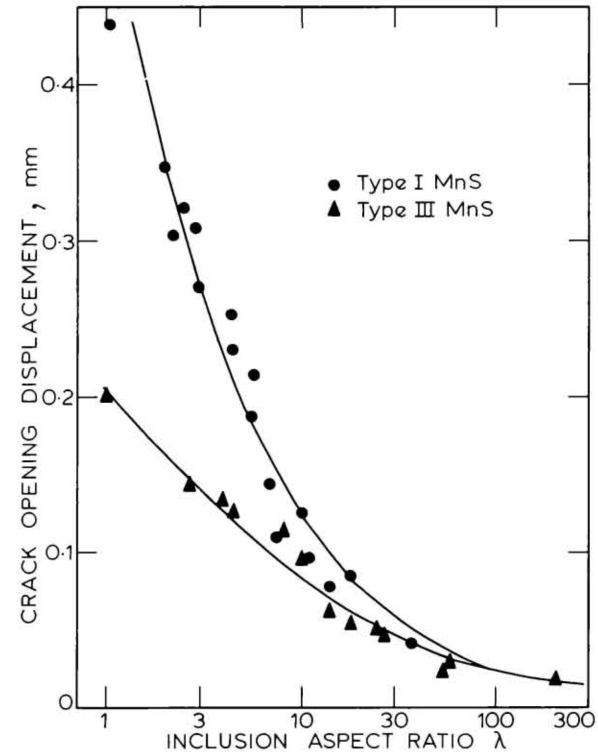
In commercial steels the final inclusion distribution is derived from a combination of deformation and reorientation phenomena, with the contribution from each being strongly dependent on the initial inclusion morphology. Thus with semikilled steels the predominant inclusions in the cast condition are globular silicates or Type I MnS, and these deform during hot working. In the case of fully killed steels the predominant inclusions are likely to be Al_2O_3 and Type II MnS. Both occur as extensive arrays and hence the major effect of hot working is to reorientate the inclusions. The influence of each of these effects on fracture behaviour will be examined in greater detail below.

Inclusion deformation

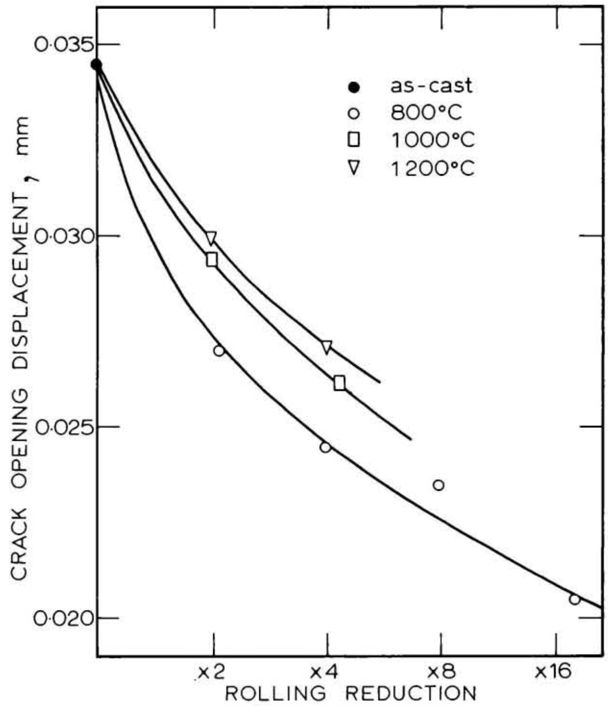
The effect of inclusion deformation on the short-transverse toughness of hot-rolled steels has been studied in two steels containing Type I and Type III MnS respectively. Both of these sulphides types have globular or equiaxed forms in the cast steel and deform to ellipsoids on hot rolling. In the steels examined, the sulphides were initially relatively large (mean diameters of 30 μm and 7 μm for the Type I and Type III respectively) and hence the final distributions were amenable to quantitative analysis.

Each steel was hot rolled under carefully controlled conditions to reductions of between ×2 and ×16 at various rolling temperatures between 1 200° and 800°C. As discussed above, the relative plasticity of the Type I MnS inclusions is low at high rolling temperatures and increases to a maximum at about 900°C when mean aspect ratios of about 40 were produced after the ×16 rolling reduction. The Type III MnS are much more highly deformable than the Type I at all rolling temperatures and at 800°C, when the relative plasticity was equal to unity, mean aspect ratios in excess of 200 were achieved. In this way a very wide range of inclusion aspect ratios was generated with frequently similar aspect ratios being developed from quite different combinations of sulphide type, rolling temperature, and reduction. After rolling, the steels were normalized and then short-transverse (S–L orientation), single-edge notched beam specimens were prepared for COD testing. The technique for testing thin strip in this orientation is discussed in detail below.[26]

The COD results for both the Type I and Type III MnS systems are plotted against aspect ratio in Fig.9. This demonstrates the very powerful influence of inclusion deformation on toughness and also shows that the effect of aspect ratio is not influenced by the particular combination of rolling temperature and reduction by which it is achieved. The results from both the Type I and Type III MnS show similar trends but the curves are not coincident owing to differences in the size and volume fraction of the inclusions. Although, as with volume fraction, the inclusion aspect ratio alone is not sufficient to characterize the effect of inclusions on fracture, a low ratio is a very necessary requisite for optimum toughness.



9 Effect of inclusion aspect ratio on short-transverse toughness



10 Effect of rolling temperature and reduction on short-transverse toughness of steel containing Type II MnS

Inclusion distribution and reorientation

The main types of non-metallic inclusion which occur as extensive arrays in cast steel are alumina and Type II MnS. Of these, the Type II MnS is probably the most serious problem since it is precipitated as extensive arrays of very fine rods in an interdendritic eutectic distribution. Moreover, MnS is usually present in this form in fully killed steels. One of the consequences of this particular morphology is that it is not possible to obtain any significant measure of the inclusion deformability or distribution from examination of polished sections since individual rods cannot be sectioned along their length, and it is not always possible to resolve the highly elongated sulphides in the optical microscope. Accordingly, these inclusions can be characterized in detail only by fractographic techniques.

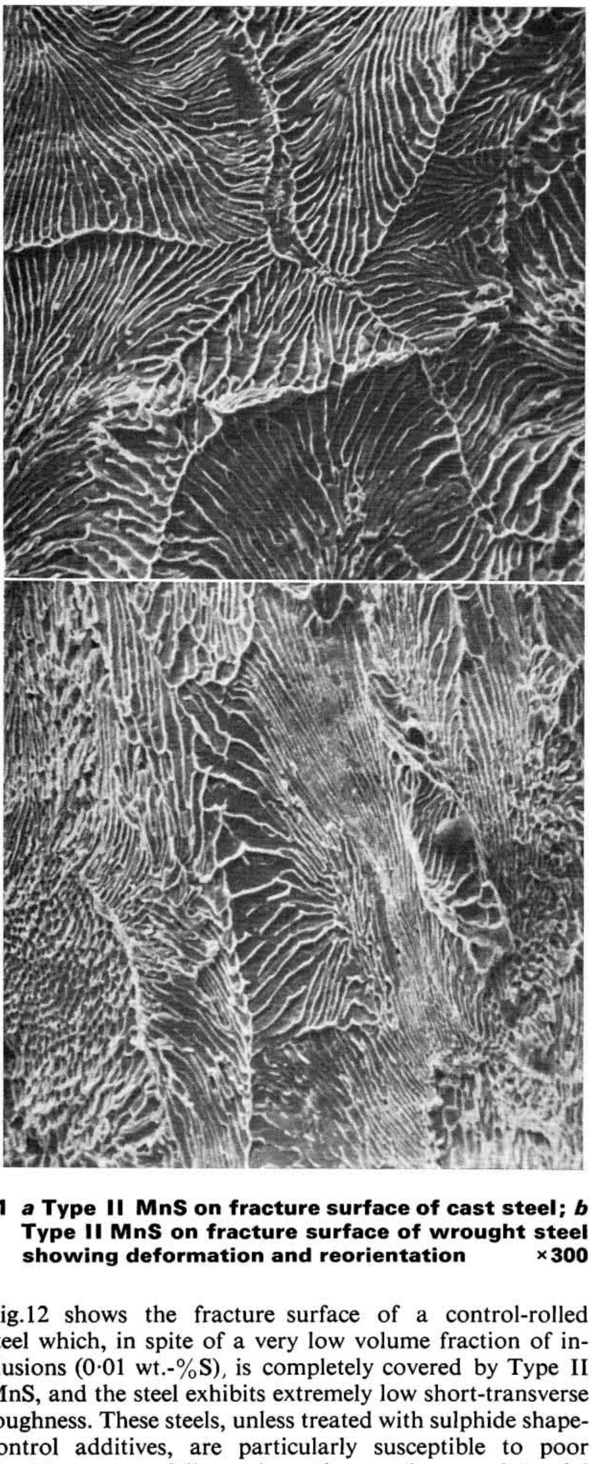
In order to study the role of Type II MnS in the fracture behaviour of hot-rolled steel, samples of cast steel have been hot rolled and tested in a manner similar to the steels containing Type I and Type III MnS. The short-transverse crack opening displacement results from the specimens rolled at each temperature are plotted against rolling reduction in Fig.10 and it is apparent that the toughness decreases with both increasing rolling reduction and decreasing rolling temperature. As expected, maximum toughness is exhibited by the cast material, but the COD at fracture was less than the minimum value shown by a similar steel containing the most highly elongated Type I MnS. Examination of the corresponding fracture surface (Fig.11*a*) shows that the interdendritic sulphide colonies provide an easy and continuous crack path through the specimen giving rise to an intergranular type of fracture. The separation between individual rods and colonies is typically 3 μm and therefore a very small amount of steel matrix is involved in the joining of microvoids.

In the wrought steels two main effects are apparent which become more pronounced with increasing rolling reductions (Fig.11*b*):

(i) the complete colonies become reorientated parallel to the rolling plane
(ii) the individual sulphide rods are elongated and flattened.

In the specimens rolled at 1200°C there is very little elongation of the sulphides and the observed effect on toughness must arise from reorientation of the sulphide colonies. This has the effect of making the fracture surfaces much flatter and it has been calculated that after a rolling reduction of ×4 the average inclination of the sulphide colonies to the macroscopic crack plane is about 7°. Thus, the crack path length is within 1% of the minimum straight-through position. This corresponds to a 20% reduction in path length from the as-cast condition which is in close agreement with the observed reduction in toughness. At lower rolling temperatures elongation of the sulphides occurs, and, as with the other sulphide types, this becomes more pronounced with decreasing rolling temperature. This produces a more rapid rotation into the rolling plane, and less matrix rupture is involved in a given length of crack, thus further decreasing the toughness. An additional effect of the deformation is that in the same way that the complete colonies become aligned with the rolling plane, the constituent arms are rotated into the rolling direction which produces extensive planes containing many parallel fibres. Moreover, because of the deformation and crowding together of the rods, their separation becomes extremely small and this is a major factor controlling the ease of crack propagation.

The maximum deformation of the Type II MnS was observed after the ×16 reduction at 800°C. This was associated with the lowest toughness and the fracture surface was covered by overlapping layers of highly elongated stringers which were strongly aligned with the rolling direction. The sulphur content of the experimental steel was 0·28%, which is very much higher than that encountered in structural steels. However, identical fracture appearances are encountered in commercial steels and a



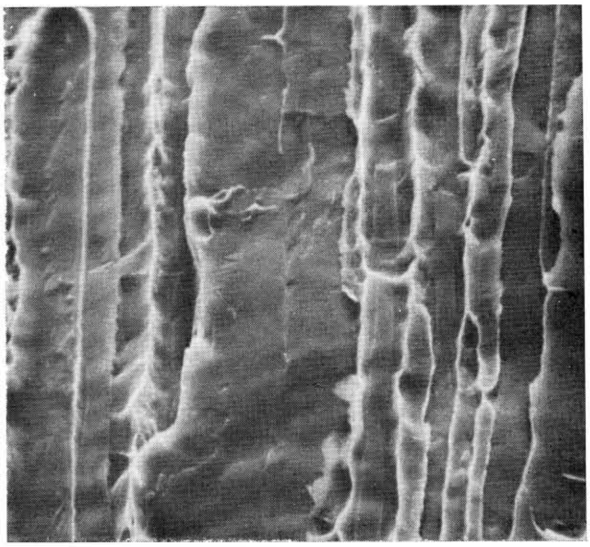
b

11 *a* Type II MnS on fracture surface of cast steel; *b* Type II MnS on fracture surface of wrought steel showing deformation and reorientation ×300

Fig.12 shows the fracture surface of a control-rolled steel which, in spite of a very low volume fraction of inclusions (0·01 wt.-%S), is completely covered by Type II MnS, and the steel exhibits extremely low short-transverse toughness. These steels, unless treated with sulphide shape-control additives, are particularly susceptible to poor short-transverse failure since they undergo substantial reductions at low rolling temperatures where the relative plasticity of the sulphides is very high. It is also important to note that in these steels the very detrimental distribution of inclusions cannot usually be identified by optical metallography owing to the difficulty of resolving the very fine sulphides and the inability to section them along their length. This has important implications regarding the assessment of steel cleanliness by metallographic techniques.

INCLUSION SIZE AND NUMBER

The number of non-metallic inclusions per unit volume,



12 Fracture surface of commercial control-rolled steel covered with Type II MnS ×2000

Nv, their volume fraction, f, and their size (semi-axes a, b, c) are related by the expression:

$$Nv = 3f/4\pi abc$$

A paradoxical situation arises when it is generally considered that for optimum toughness the number of inclusions should be as low as possible while their size should be as small as possible. Clearly these two requirements are incompatible at constant volume fraction. It is well established that steel refining and casting techniques such as electroslag remelting and vacuum arc refining produce small inclusions and are usually associated with high toughness. However, it is difficult to ascribe the improved toughness to the reduced inclusion size since there are invariably simultaneous changes in the inclusion shape and volume fraction, both of which have already been shown to have a powerful effect on toughness. As discussed earlier, the small inclusions are most resistant to deformation owing to their high surface/volume ratio, and it is found that the 1–2 μm size sulphide inclusions, which are encountered in ESR and VAR steels, undergo negligible deformation even after substantial hot-working reductions.

Surprisingly, little work has been carried out on the effect of inclusion size on toughness. From the point of view of void initiation, Ashby's dislocation model[27] predicts an inverse relationship between inclusion size and the stress required for void initiation, and this tends to be confirmed by Palmer's observation of voids associated with very small SiO_2 particles in copper.[28] On the other hand, the Brown and Stobbs energy criterion[29] predicts a very weak effect of particle size on void initiation with voids forming preferentially at small particles. In the case of large inclusions in steel the situation is by no means clearly defined. Roesch[30] in his studies of alumina in steel observed no effect of size within the range 0·05–40 μm but these experiments utilized sintered specimens and hence must again be interpreted with caution. In the authors' own work on sulphides, voids were observed simultaneously on inclusions ranging in size from 3 to 30 μm. The voids were larger around the larger inclusions but the ratio of void size to inclusion size remained almost constant.

In order to achieve fracture, both void initiation and coalescence are required. Since the spacing of the inclusions is directly proportional to their size, it might be expected that fine inclusions would be more detrimental than large ones. In our investigations of steels containing Type I and Type III MnS it was found that for similar aspect ratios the toughness of the steel containing 0·4 vol.-% of Type III MnS with a mean size of 7 μm was in fact inferior to the steel containing 1·4 vol.-% of Type I MnS with a mean size of 30 μm. The matrix properties of the two steels were somewhat different but the results do suggest that smaller inclusions are associated with reduced toughness and, as will be discussed subsequently, this does correlate with the reduced interinclusion spacing. This is also supported by the work of Joy and Nutting[31] on overheated steels where a correlation is established between inclusion size, spacing, and toughness.

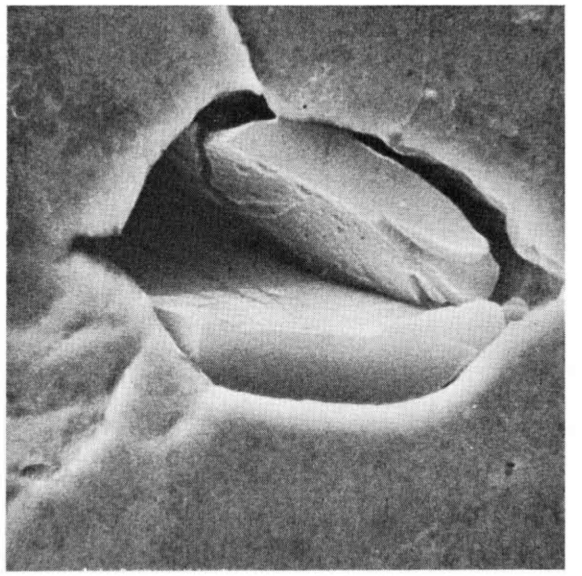
Work on overheated steels does suggest, however, that there may be a limit to which reduced inclusion size is reflected in reduced toughness. In overheated VAR steels containing low volume fractions of MnS the individual sulphides and their spacings are extremely small (typically 0·5 and 2·0 μm respectively), and yet this is not reflected in any significant reduction in toughness.[32] This is despite the fact that the volume fraction of sulphides in the zone of fracture is of the order of 1 vol.-%. This is similar to the volume fraction of sulphides in a free-cutting steel, where, with large inclusions, very low toughness is observed.

In summary, it can be said that the effect of inclusion size on toughness is not yet clearly established. Present evidence suggests that toughness decreases with decreasing size down to perhaps 3–5 μm. Thereafter, the effect of the inclusions becomes less detrimental. In practice, however, the beneficial effect of reduced inclusion deformation with reduced size is likely to outweigh any inherent effect of size on toughness.

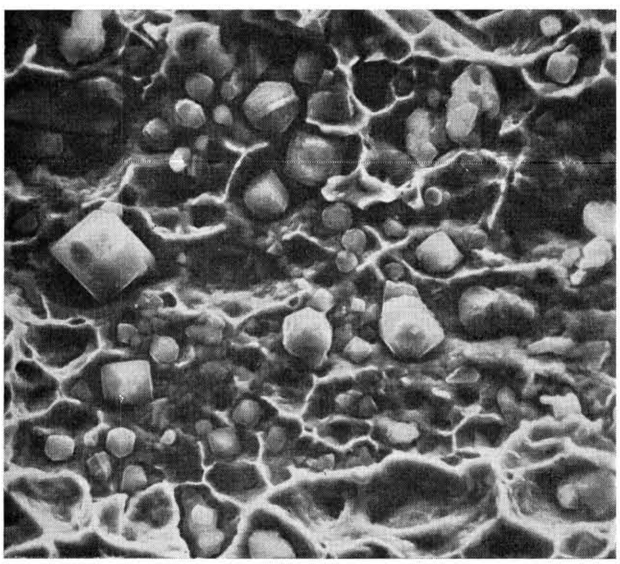
VOID FORMATION DURING ROLLING

When the inclusion phase is harder than the matrix, voids and discontinuities extending from the inclusion into the matrix may form during rolling (Fig.6). It has been suggested that these are produced only if the relative plasticity of the inclusion is less than 0·5,[33] and this is confirmed by the fact that they are commonly associated with silicate inclusions rolled at low temperature, and with Type I MnS only when rolled at temperatures in excess of 1200°C. Their formation appears to be due to the very low strength of the inclusion/matrix interface which is insufficient to withstand the longitudinal tensile stresses which are set up by the flow of steel around the non-deforming inclusions.

In most cases the conical voids are observed to extend into the matrix for a distance equal to about half the inclusion height. Occasionally, very much longer discontinuities are observed associated with silicate inclusions of low deformability. Here, the total length of discontinuity associated with any inclusion is normally equal, approximately, to the initial inclusion diameter multiplied by the rolling reduction and results from a failure of the flowing steel interfaces leaving the inclusion to weld together. This failure to weld is probably due to the presence of a thin film of MnS originally present as a shell surrounding the silicate inclusion. The matrix ahead of the voids has not normally been found to provide an easy path for crack propagation (Fig.13) and the main effect of the voids appears to be simply to increase the effective length of the inclusions. This is a relatively minor effect compared with inclusion deformation. The importance of this phenomenon is that if an inclusion fractures during deformation, then the fragmented parts must become separated by a considerable distance before internal welding of the matrix can take place between them. Although the extent of voids associated with non-deforming spherical inclusions is typically similar to their diameter, the flow of steel around fragmented parts is expected to be more unfavourable for internal welding and it may be necessary to achieve a separation of several times their size before the continuous defect is eliminated. The inclusions giving rise to defects of this type are predominantly alumina, titanium carbosulphides and silicates, and Fig.14 shows part of a very large colony of fragmented alumina particles in which there is negligible matrix bonding between the fragments. In such cases the total effective inclusion size must be considered to be equal to the maximum extent of the fragments.



13 Inclusion adjacent to fracture surface; void linkage by localized shear mechanism indicates absence of any matrix embrittlement ahead of the conical voids developed during rolling ×2500



14 Alumina colony showing absence of matrix bonding between fragments ×800

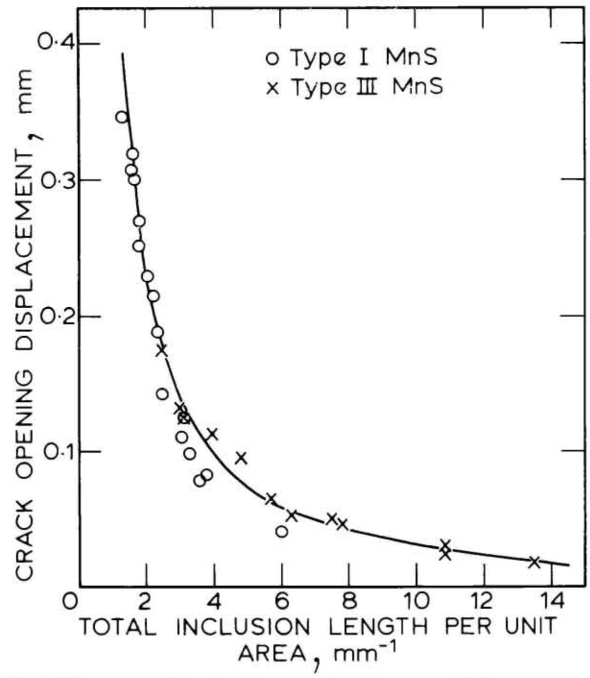
Inclusion parameters for assessing the fracture behaviour of steel

The preceding sections have established that inclusion volume fraction, shape, and size all contribute to the effect of inclusions on toughness. Any single parameter for assessing the effect of non-metallic inclusions on steel toughness must accordingly take all these factors into consideration. One such parameter which does show considerable promise is the total projected length of inclusions per unit area. In a random array of elliptical inclusions with semi-axes a, b, c and a total volume fraction of f, the total inclusion projection on a section normal to the c-axis and in the direction of a, is given by:

$$P=\frac{2f\lambda}{3\pi a}$$

where λ is the aspect ratio. This parameter incorporates all the inclusion variables, and it has the additional advantage that it is convenient to measure and is amenable to automatic inclusion-counting techniques. Moreover, in the case of highly elongated inclusions it is frequently the only parameter which can be measured with any accuracy.

For the various deformed Type I and Type III MnS systems discussed previously, the total projected inclusion length per unit area has been measured on transverse sections for each specimen. These values are plotted against the short-transverse toughness results in Fig.15 and it is apparent that despite large differences in volume fraction, aspect ratio, and size, the results from both steels now fall on a single curve. This projection parameter has also been found to provide the best correlation with the short-transverse toughness of a wide range of steels which are susceptible to lamellar tearing[34] and a modification of this parameter has been found to provide a useful correlation with the transverse upper shelf energy of line-pipe steels.[35]



15 Influence of inclusion projection on COD

Mechanism of short-transverse fibrous fracture

The basic features of fibrous fracture are now well established. Thus it has been shown that void formation occurs at non-metallic inclusions at very low plastic strains and subsequently, due to strain concentration at these voids, they grow and eventually coalesce. It is usually considered that coalescence occurs by the lateral expansion of the voids with associated necking down of the intervening matrix. On the basis of this mechanism several attempts have been made to develop mathematical models which relate the toughness of the steels to the inclusion dispersion.[21–24, 36–38] In all these models it is considered that final fracture occurs by a necking down of the intervening matrix and the critical condition for fracture instability is considered to be either that the extent of lateral void growth becomes equal to the particle spacing or that the ratio between the extent of void growth in the direction of applied stress and the particle spacing in a direction normal to the applied stress achieves some critical value. Thus Brown[38] considers that this ratio is equal to unity, at which point 45° slip can be accommodated in the ligament and, because the restraining influence of the surrounding matrix is no longer present, necking down of the ligament can occur as in a tensile specimen. Some success has been achieved with these models, principally by Gladman *et al.*,[23] but unfortunately they are not applicable in the case of short-transverse fractures since void coalescence does not occur by a necking down of the intervening matrix. Instead, inclusion linkage occurs here by a highly localized shear mechanism with negligible lateral expansion of the voids as shown in Fig.16. The effect of increasing inclusion deformation is that the spacing between adjacent inclusions in the direction of fracture rapidly decreases and the region of localized shear becomes progressively more localized and orientated towards the direction of applied stress. In all cases the lateral void growth is negligible whereas the extent of void growth in the direction of applied stress decreases with decreasing void spacing. Immediately before linkage, the ratio between the void size and the inclusion width in the direction of applied stress is remarkably constant at a value of about 2. Since the inclusion spacing is proportional to the inclusion size, this implies that there is a constant ratio between void size and spacing at fracture. However, since linkage does not occur by continued void expansion, it is considered that this merely reflects the local strain required to achieve the conditions necessary for unstable shear fracture.

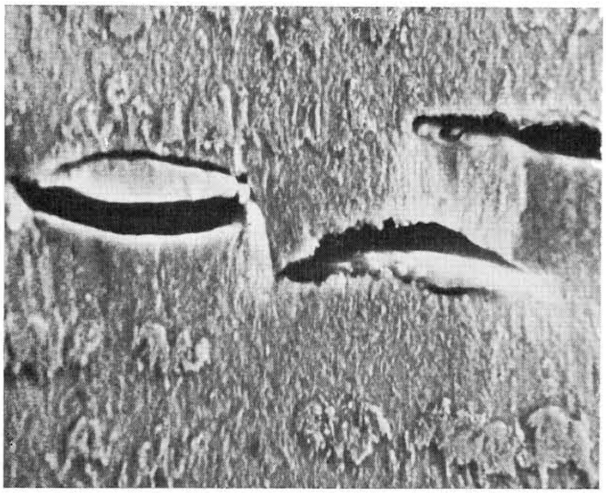
The most pronounced effect of increased deformation is that the ratio of short-transverse to longitudinal separation decreases rapidly. This accounts for the fractographic observation that as deformation increases the crack tends to increase progressively by a series of steps in the transverse direction. The most favourable direction for inclusion joining must be a compromise between the direction of minimum separation and the direction of greatest ease of matrix fracture. This direction has been shown[26] to be inclined at an angle of $\tan^{-1}\sqrt{1-2/\lambda^2}$ to the longitudinal direction and the inclusion separation is then given by:

$$d=\frac{\sqrt{2a(1-f)}}{3\lambda f}$$

This expression has a very similar form to that for the projection parameter P and for low inclusion volume fractions it may be shown that:

$$d\simeq\frac{2\sqrt{2}}{9\pi P}$$

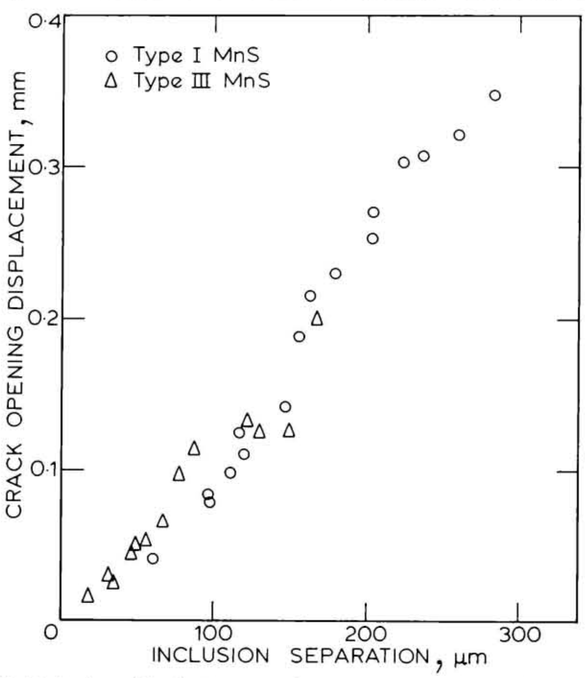
The projection parameter is therefore inversely proportional to the interinclusion spacing in the direction of



16 Section through arrested crack showing shear linkage of inclusions ×160

fracture. The previous toughness data may therefore be replotted against inclusion spacing and, as shown in Fig.17, this reveals a linear relationship for both the inclusion systems studied. A similar relationship between short-transverse toughness and inclusion spacing has been observed in the case of forged aluminium alloys[39] and it may be concluded that the inclusions determine the toughness simply by controlling the volume of the zone of highly strained matrix which is involved in the fracture process.

Inclusion spacing, therefore, adequately accounts for the effect of hot working on short-transverse toughness. However, this takes no account of the matrix strength and work-hardening characteristics and is not true for all inclusion dispersions and orientations. Thus, by employing high-temperature homogenization treatments on steels containing highly elongated sulphide inclusions it is possible to spheroidize the inclusions and thereby obtain substantial improvements in toughness but without any significant reduction in inclusion spacing.[20] Likewise, by



17 Relationship between short-transverse toughness and inclusion separation in direction of fracture

changing the orientation of the test specimen so that the elongated inclusions are parallel to the direction of applied stress, a very large increase in toughness is obtained while the inclusion spacing obviously remains constant. As discussed above, this orientation is associated with a different mechanism of void linkage, that is, internal necking rather than localized shear. Consequently, it is necessary to consider the factors controlling the onset of shear instability in rather more detail. One of the few attempts to analyse this situation has been made by McClintock,[37] who found that his earlier homogeneous void coalescence models predicted much higher ductilities than were observed in practice, and a simplified version of his condition for shear instability between inclusions is as follows:

$$\frac{1}{\sigma}\left|\frac{d\sigma}{d\epsilon}\right| < kF^2\left(\frac{f}{1-f}\right)^2\sqrt{\lambda^2+1}$$

where k is a constant, F is a hole growth factor, f is the volume fraction of inclusions, λ is the aspect ratio of the voids at instability (as discussed above, this is observed to be about half the aspect ratio of the original inclusion), $d\sigma/d\epsilon$ is the rate of work hardening at instability.

The model is not quite correct since it considers a shear band encompassing the complete inclusion rather than the very localized shear zones which have been observed. However, the implications are obviously correct and the very powerful effect of volume fraction and aspect ratio is again clearly demonstrated. The effects of inclusion spheroidization and reorientation may, therefore, now be accounted for in terms of their greater resistance to shear instability. The McClintock instability condition also clearly demonstrates that, in addition to the effect of inclusions, the resistance to shear instability, and thereby the toughness, is strongly influenced by the strength of the matrix and the rate of work hardening. It is important to note that this is the work-hardening rate at very high strains and the improvements in toughness which can be obtained by increasing this are clearly demonstrated by the outstanding toughness exhibited by the TRIP class of steels.[40]

In conclusion, the major role of inclusions in short-transverse fractures is to produce strain localization and hence establish the conditions necessary for shear instability in the matrix at much lower strains than would be necessary in their absence, with the magnitude of the overall strain being controlled by the interinclusion spacing in the direction of shear linkage.

Conclusions

The practical implications arising from the foregoing discussion are that in order to achieve maximum toughness and minimum anisotropy in hot-rolled steels the following objectives should be sought:

(i) the volume fraction of inclusions should be as low as possible

(ii) the inclusions should be uniformly dispersed throughout the steel

(iii) the inclusions should have a compact morphology in the cast steel

(iv) the hardness of the inclusions should preferably be double that of the steel during hot working so as to minimize their deformation.

In recent years there has been intense activity in the pursuit of these objectives and outstanding progress has been made in reducing the detrimental effects of non-metallic inclusions. Control of sulphides has received the greatest attention, improvements being achieved at all stages in the steelmaking process. More effective desulphurization of blast-furnace iron is now being achieved by treatment with calcium carbide, calcium cyanamide, lime, and more recently, magnesium. During steelmaking the effectiveness of slag–metal reactions has been improved by means of more efficient stirring and injection techniques, and the addition of strong sulphide formers such as calcium and rare earths to the ladle has resulted in further reductions in sulphur content. Also, remelting techniques such as electroslag refining permit further reductions in sulphur by the use of lime-rich 'active' slags. By means of these techniques, sulphur contents less than 0·01% and in some cases lower than 0·005% are now being achieved.

Having reduced the total sulphur content, it is important to ensure that the residual sulphur is not precipitated in the harmful Type II morphology and is sufficiently hard to resist deformation during working. This is now being achieved by the addition of strong sulphide-forming elements such as Ti, Zr, Ca, and rare earths. These alloy additions are expensive and complications can arise from the interaction of Ti and Zr with carbon and nitrogen. Also, the high affinity of Ca and the rare earths for oxygen calls for sophisticated addition and teeming techniques if good recoveries are to be achieved. Nevertheless, provided sufficient care is taken and appropriate amounts of alloying additions are made, toughness anisotropy can now be practically eliminated.[43-45]

Regarding the control of oxide inclusions, the increasing use of vacuum degassing techniques permits more efficient deoxidation and the incorporation of extended holding and stirring of the molten steel facilitates removal of the primary oxides. Remelting techniques such as electroslag and vacuum arc refining are also extremely effective in reducing the number and size of oxides in premium-quality steels. As with the sulphides, it is still important to ensure that the residual oxygen is present in the least deleterious form. This can again be achieved by calcium and rare-earth treatment, the deformable silicates and alumina colonies being replaced, ideally, by widely dispersed globular calcium aluminates and rare-earth oxides which do not deform during hot working.

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