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# An examination of diffusion bonded interfaces in a mild steel

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A study of the interface surfaces of fractured diffusion bonds in a mild steel was made in order to further understanding of the mechanisms of bond formation and of the mechanical properties of such bonds. The results show that parent-metal tensile strengths are achieved with some 20% of the interface still unbonded and that these unbonded areas are present as grooves rather than as isolated voids as had previously been thought. The relative orientation during bonding of any directional surface finish on the two component halves affects the geometry of the unbonded grooves and, to a slight extent, the resultant tensile strengths. The results also suggest that tensile testing alone is not a sufficient criterion by which bond quality should be assessed; impact or fracture toughness testing also must be considered.

## 1. Introduction

Diffusion bonding is a joining process in which two materials are brought together under heat and pressure, but without macroscopic deformation, for a time sufficient to allow a bond to be formed between them. Although theoretical and experimental studies, for example [1–3], have been undertaken in attempts to understand the process, a general account of all possible bonding mechanisms involved has not yet been formulated.

The present work is another contribution to the research on the diffusion bonding of metals. It differs somewhat from its predecessors in that it takes as its starting point the direct observations of (fractured) interface surfaces. An attempt is then made to interpret such observations in terms both of the mechanisms of bonding and of the mechanical properties of diffusion bonds. It is hoped that this direct observational approach will lead not only to a general appreciation of the dominant mechanisms (particularly those involved in the early stages), but also will facilitate the development of a model, capable eventually of predicting all the stages and conditions necessary for the formation of a bond.

## 2. Experimental procedure

### 2.1. Material

The material chosen for study was cold drawn mild steel (12.5 mm diameter En3B, to BS 970 1955), chosen because it gave a direct comparison with earlier work in both diffusion bonding [4] and friction welding [5]. The two rods used in this work were individually analysed (Table I) and note was taken of their use for individual specimens because Stockham [4] had shown that the parent-metal properties and bonding characteristics of En3B are very sensitive to minor variations in composition. Mild steel (not analysed but nominally En2, composition as indicated in Table I) in the form of 0.05 mm thick shim was used in some trials as an interlayer between the rods to be bonded.

### 2.2. Equipment

All bonds were made in a small diffusion bonding unit which consists of a vacuum chamber (approximately $10^{-4}$ to $10^{-5}$ torr) with a high frequency induction heater and a simple coil spring for applying pressure to the rods; further details can be found in [6]. Temperature was measured by

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TABLE I Composition of materials used

| | Element | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| | C | S | P | Si | Mn | Ni | Cr | Al | Nb | Cu |
| *Analyses of En3B* | | | | | | | | | | |
| Rod 1 | 0.12 | 0.033 | < 0.055 | 0.03 | 0.62 | < 0.01 | < 0.01 | < 0.005 | < 0.005 | < 0.01 |
| Rod 2 | 0.13 | 0.020 | 0.015 | 0.04 | 0.44 | 0.01 | < 0.01 | < 0.005 | < 0.005 | 0.02 |
| *Specification of En2* | | | | | | | | | | |
| shim | 0.20 | < 0.06 | < 0.06 | 0.10 | 0.80 | N/A | N/A | N/A | N/A | N/A |

N/A = not available

spot-welding a chromel/alumel thermocouple to the specimen surface close to the interface.

## 2.3. Preparation and bonding

Each specimen was cut to a length of 40 mm and the ends were ground square to the axis to a finish better than 0.4 $\mu$m CLA. Immediately prior to bonding, each rod was degreased with acetone. Two rods were then loaded into the chamber with the faces to be bonded in contact and the pre-set load was applied in the cold. The full setting-up procedure has been outlined by Stockham [4]. In some trials, however, the bonding faces were first heated in vacuum for 60 min before being assembled as above for bonding. In other trials, foil interlayers of steel shim were placed between the bonding faces – in these cases the shim was scratch-brushed with degreased steel wool and washed with acetone before assembly.

The bonding conditions used in all trials were 7 N mm$^{-2}$ pre-set pressure, 900° C and times varying from 2 to 120 min. These were the conditions shown by Stockham [4] to give rise to a wide range of bond strengths, thus allowing some of the various stages of bonding to be distinguished, Fig. 1.

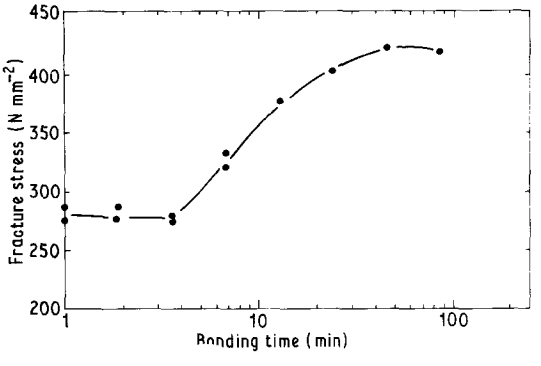



*Figure 1* Variations of fracture stress with time at a bonding temperature of 900° C and pressure of 7 N mm$^{-2}$ [4].

## 2.4. Mechanical testing and fractography

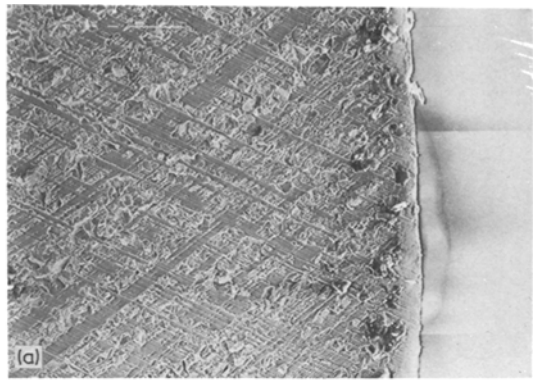
The strength of as-bonded specimens, i.e. without further machining, were measured by tensile testing the specimens to destruction. The maximum load only was recorded because misalignment, notch effects, etc, were undoubtedly present and invalidated other measurements. The reasons for using unmachined specimens were twofold: firstly to allow the whole bonded surface to be examined, and secondly, to maintain comparability with Stockham [4]. As-bonded fracture faces, undistorted by gross deformation, were obtained by longitudinally slicing the bonded pair (by spark erosion) and then fracturing one half at liquid nitrogen temperature by a hammer blow. All fracture faces were subsequently examined in a scanning electron microscope.

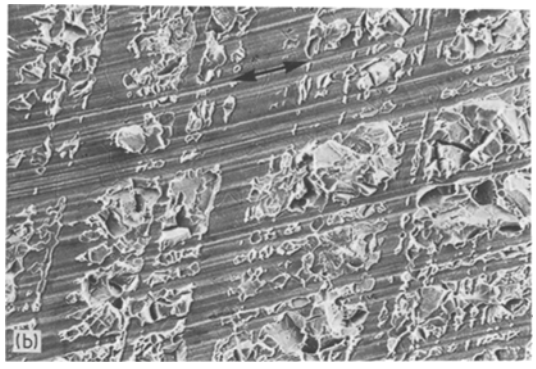
## 2.5. Effect of heat on machined surfaces

The changes which occurred during the heating of a freely suspended, pre-ground surface in vacuum were studied by heating small ground discs of En3B steel in a vacuum induction furnace at 900° C for times between 2 and 120 min. The resultant in surface topography were asserted by: (a) scanning electron microscopy; (b) surface profile measurements, using both a Talysurf with trace analysis facility (by courtesy of Leicester Polytechnic Tribology Section) and a Talystep (by courtesy of the National Physical Laboratory). An *in situ* study of free-surface heating in vacuum was also attempted in a hot-stage optical microscope.

## 2.6. Bonded area estimates from fracture surfaces

Scanning electron micrographs (at × 500) of the fracture surfaces were assessed for percentage of bonded area using a quantitative image analyser. It was necessary to make "block tracings" around what were judged to be the bonded regions in order to provide sufficient contrast for instrument



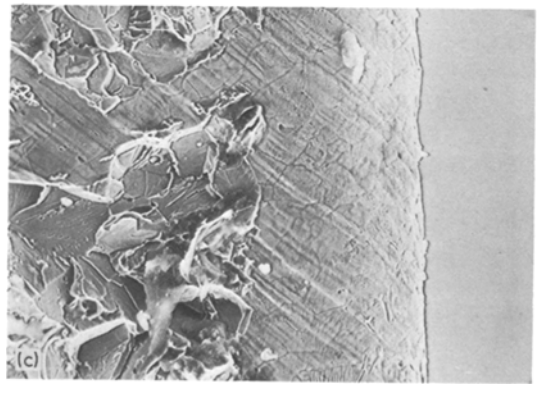




*Figure 2* SEM photographs of fracture surfaces. En3B bonded for 15 min at 900° C and 7 N mm$^{-2}$, fractured at low temperature: (a) General view, edge of specimen, × 72; (b) Same area, × 290; (c) Edge of specimen, × 580.

discrimination; tracings from three randomly chosen areas were taken from each fracture face.

Other methods of estimating bonded area were found to be unsuitable (e.g. direct analysis using an image analyser of the fracture surface itself or of a photograph, or direct point counting from the fracture surface). Various techniques such as etching and oxidation were attempted in order to improve the contrast between bonded and unbonded areas, but they too were ineffective.

## 3. Results and interpretation

### 3.1. Scanning electron microscopy of fracture surfaces

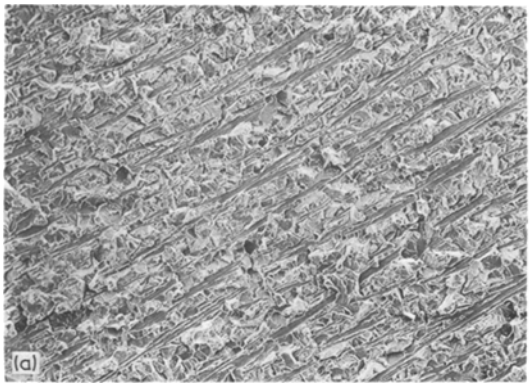
Figs 2 and 3 are typical of the surfaces resulting from liquid nitrogen impact fractures of bonds made with random orientation of grinding marks. Fig. 2a is a low magnification of the fracture surface of a specimen bonded for 15 min under standard conditions; it includes the edge of the sample. Fig. 2b shows the cross-hatched markings in more detail, and the edge of the sample is shown at higher magnification in Fig. 2c. The photographs in Fig. 3 are from a specimen bonded for 120 min. They show that the same features are still present, but have developed in detail.

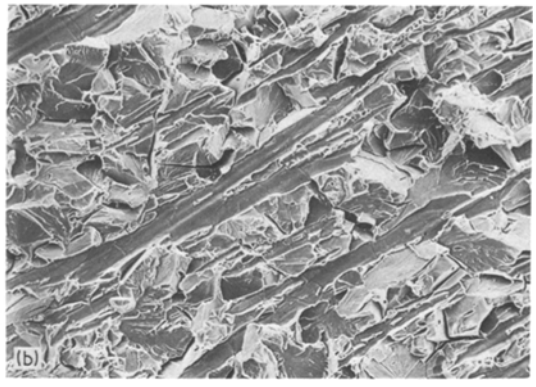
Several features of a bond interface become apparent from examination of these figures.

(i) The geometry of the non-bonded areas is controlled by the original orientation of the machining marks on the two mating surfaces. Thus, in Fig. 2, the machining marks were, by chance, aligned almost perpendicular to one another so giving the cross-hatched appearance shown . In addition, the continuous machining marks arrowed in Fig. 2b show that the "horizontal" (as printed) unbonded areas originate from grooves in the observed surfaces; it follows that the (as printed) "vertical" unbonded areas are produced by virtue of being opposite grooves on the other face. In the specimen shown in Fig. 3, however, the grinding marks on each surface were aligned almost parallel to one another, thus giving rise to a different bonding pattern. The significance of these patterns in terms of bond strengths will be discussed later (Section 3.3.).

(ii) An examination of the unbonded areas shows that thermal etching or surface diffusion has resulted in grain boundary grooving. This is illustrated very clearly, see Fig. 3b, from a specimen bonded for 120 min and also is especially apparent at all specimen edges, Figs 2c and 3c. The phenomenon indicates that all unbonded areas act as free surfaces throughout the bonding process.

(iii) An "edge effect" (Figs 2c and 3c) is apparent on all cold-fractured surfaces examined; it constitutes a notch and must therefore have a significant effect on the subsequent mechanical testing of unmachined specimens. It is quite possible that such a notch exists in all low-



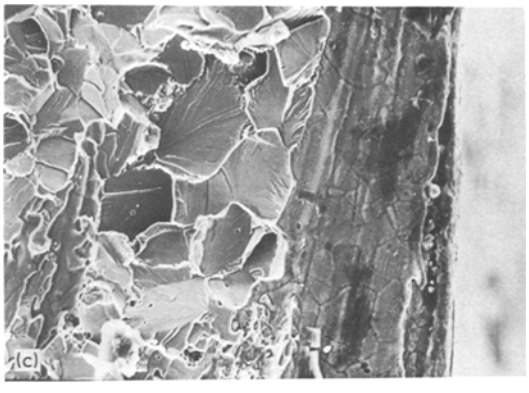




*Figure 3* SEM photographs of fracture surfaces. En3B bonded for 120 min at 900° C and 7 N mm$^{-2}$, fractured at low temperature: (a) General view, × 72; (b) Same area, note extensive grain boundary grooving, × 290; (c) Edge of specimen, × 580.

deformation diffusion bonded components. It is unlikely to be produced by "bounce" in the grinding machine because it appears as a complete ring of uniform width all round the edge of the surface, i.e. it does not appear simply as a sloping termination to the grinding marks. The effect could be produced by the release of residual surface stresses and it may be enhanced by the ease of surface diffusion at an edge.

(iv) Polished cross-sections of diffusion bonds create the impression that the final distribution and shape of the unbonded regions is a row of more or less elliptical voids, Fig. 4. However, the present observations of fracture surfaces imply that such "voids" are more likely to be tunnels stemming from the original machining marks. The difference between these two geometries will be important both when considering the mechanisms responsible for bonding (e.g. minimizing interfacial voids) and when interpreting mechanical properties.

These various observations led to three further areas of study: (a) the heating, in vacuum, of machined surfaces in order to simulate the changes which occur in the unbonded regions at an interface; (b) the effect of grinding mark orientation on mechanical properties; (c) the correlation of bonded area percentage with mechanical strength and orientation effects.

### 3.2. Examination of heated free surfaces

The effects of heating free (ground) surfaces in vacuum at 900° C are shown in Fig. 5. Fig. 5a is

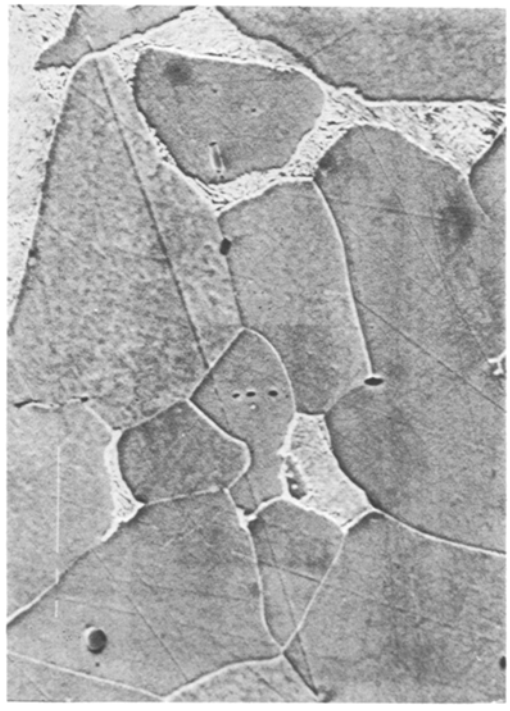


*Figure 4* Cross-section showing interface voids through En3B specimen bonded for 20 min at 900° C and 7 N mm$^{-2}$, × 1560.

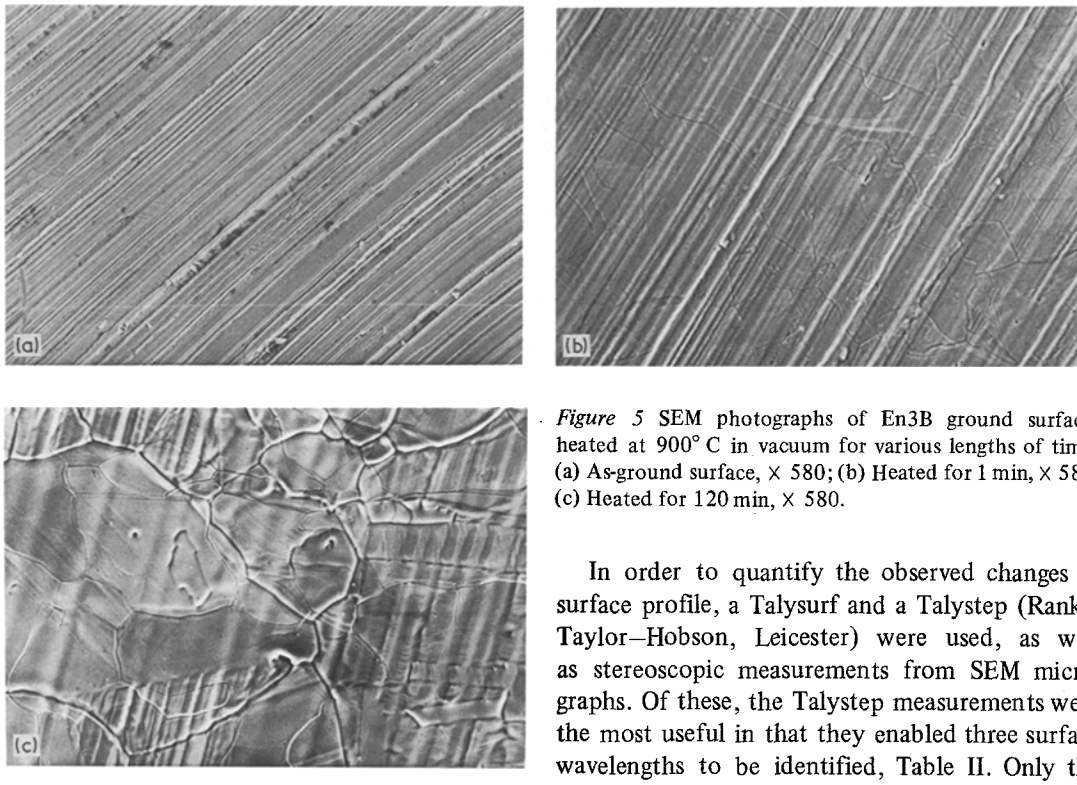


*Figure 5* SEM photographs of En3B ground surfaces heated at 900° C in vacuum for various lengths of time: (a) As-ground surface, × 580; (b) Heated for 1 min, × 580; (c) Heated for 120 min, × 580.

a scanning electron micrograph of the initial as-ground surface (surface roughness of 0.17 μm CLA); Fig. 5b shows that thermal grooving develops very rapidly and can be seen after heating for only 1 min. After a total heating time of 120 min at 900° C the surface appearance is very complex; there are both deeply grooved and lightly grooved grain boundaries, twins, possibly slip lines, and smooth areas which have a markedly crystallographic delineation.

Continuous observation of heated surfaces in a hot-stage optical microscope reveals that the deeply grooved boundaries mark the austenite grains, with some grain boundary migration giving rise to ghost boundaries, whereas the lightly etched boundaries are produced as a result of ferrite nucleation on cooling.

TABLE II Effect of heating on grinding asperities (En3B substrates)

| As-ground | | Heated at 900° C for 120 min | |
|---|---|---|---|
| Wavelength (μm) | Amplitude (μm) | Wavelength (μm) | Amplitude (μm) |
| 100 ± 15 | 0.35 | 100 ± 15 | 0.50 |
| 15 ± 3 | 0.15 | 15 ± 3 | 0.18 |
| 1.5 ± 1 | 0.015 | — | — |

In order to quantify the observed changes in surface profile, a Talysurf and a Talystep (Rank–Taylor–Hobson, Leicester) were used, as well as stereoscopic measurements from SEM micrographs. Of these, the Talystep measurements were the most useful in that they enabled three surface wavelengths to be identified, Table II. Only the shortest of these wavelengths appears to be affected by heating the free surface.

## 3.3. Tensile results

Four sets of samples were bonded at 900° C and 7 N mm$^{-2}$ for various times and were subsequently fractured in tension to determine their fracture strengths. The sets were: (a) bonds with perpendicularly aligned machining marks; (b) bonds with machining marks aligned parallel; (c) surfaces separated and preheated for 60 min at 900° C in vacuum before bonding and (d) bonds (perpendicularly aligned machining marks) with a steel shim interposed.

Specimens were tensile tested in the as-bonded state and the resultant failure stresses are presented in Fig. 6. A wide scatter of results is seen but, because only a small number of tests was performed, it is not known if the scatter is due to some mechanical effect (e.g. a notch, axial misalignment or load), or if it is inherent in the bonding process. All specimens failed at the bond line after some apparent yielding, but none reached more than approximately 90% of parent-metal tensile strength. Only the maximum fracture stress was recorded because neither yield nor proof stresses were sufficiently well defined.

The difference in fracture stresses between the

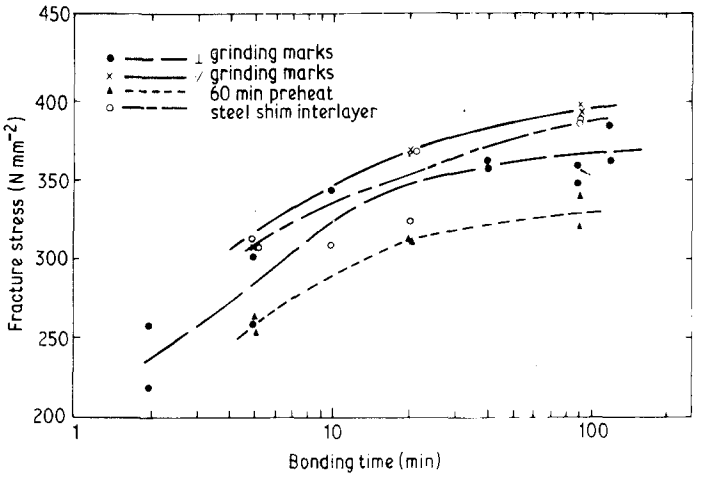



*Figure 6* Variation of fracture stress with time at a bonding temperature of 900° C and pressure of 7 N mm$^{-2}$.

four groups of specimens is less than might be expected from observation of the fracture surfaces (shown in Fig. 7). The highest fracture levels for any particular bonding time are observed when using parallel machining marks or shim interlayers. Although the differences between these two groups might be found to be insignificant if the true extent of scatter of results could be determined, there does appear to be a consistent trend which indicates a slight superiority for the parallel alignment group. Perpendicular alignment (although showing a greater scatter than all the other groups) appears to give a consistently lower fracture value, Fig. 6.

Preheated surfaces gave significantly lower values and several possible explanations are suggested.

(i) When (free) ground surfaces are preheated, the overall effect is a smoothing of the finest asperities (Section 3.2.). Thus the initial pressure-welding effect, due to deformation of the asperities, is decreased and this may lead to a possible lengthening of the bond initiation stage [1].

(ii) Preheating results in a decrease in stored energy (introduced during the surface machining operations). As this stored energy may make a significant contribution to the initial stages of bonding, its removal would be disadvantageous to bonding.

(iii) Preheating produced a certain degree of decarburization at the specimen surface and also caused some oxidation (which was not removed before bonding) at the vacuum levels employed ($10^{-3}$ to $10^{-4}$ torr). This could lead to two effects, i.e. a wide layer of ferrite at the interface and a greater amount of oxide to be dispersed before a bond could form.

Some evidence to support the third of these suggestions is described below, although this is not to imply that the first two mechanisms are not equally as important.

The fracture surfaces of bonds made between two preheated surfaces show areas of very fine ductile dimples situated at the bond interface, but brittle cleavage is present on either side of the bond where parent material has been pulled out (as shown in Figs 7c and e). Similar cleavage facets, but much coarser dimples, are observed on fracture faces from samples in which pre-bond surface heating had not been employed (see Figs 7a, b and d). The finer dimples observed on the "preheated" fracture surfaces are probably a result of microvoid coalescence around a fine dispersion of oxide particles in the ferrite layer at the bond interface. The dimples on the other fracture faces (i.e. bonds without preheated surface) appear to be formed as a result of conventional ductile failure in the ferrite layer at the bond interface, without the added void nucleation by the postulated fine oxide particles. This latter observation supports the hypothesis that the rather poor levels of vacuum employed during pre-bond surface heating resulted in a greater amount of interface oxide to be dispersed than otherwise would have been present.

Finally, it has been proposed [7] that the presence of a smooth shim between the two faces to be bonded would effectively halve the depth (and therefore the volume) of individual voids, and so lead to a significant improvement in mechanical properties. While the fracture surface, Fig. 7d, demonstrates that the shim does indeed provide a smooth surface which separates the two machined faces, the results in Fig. 6

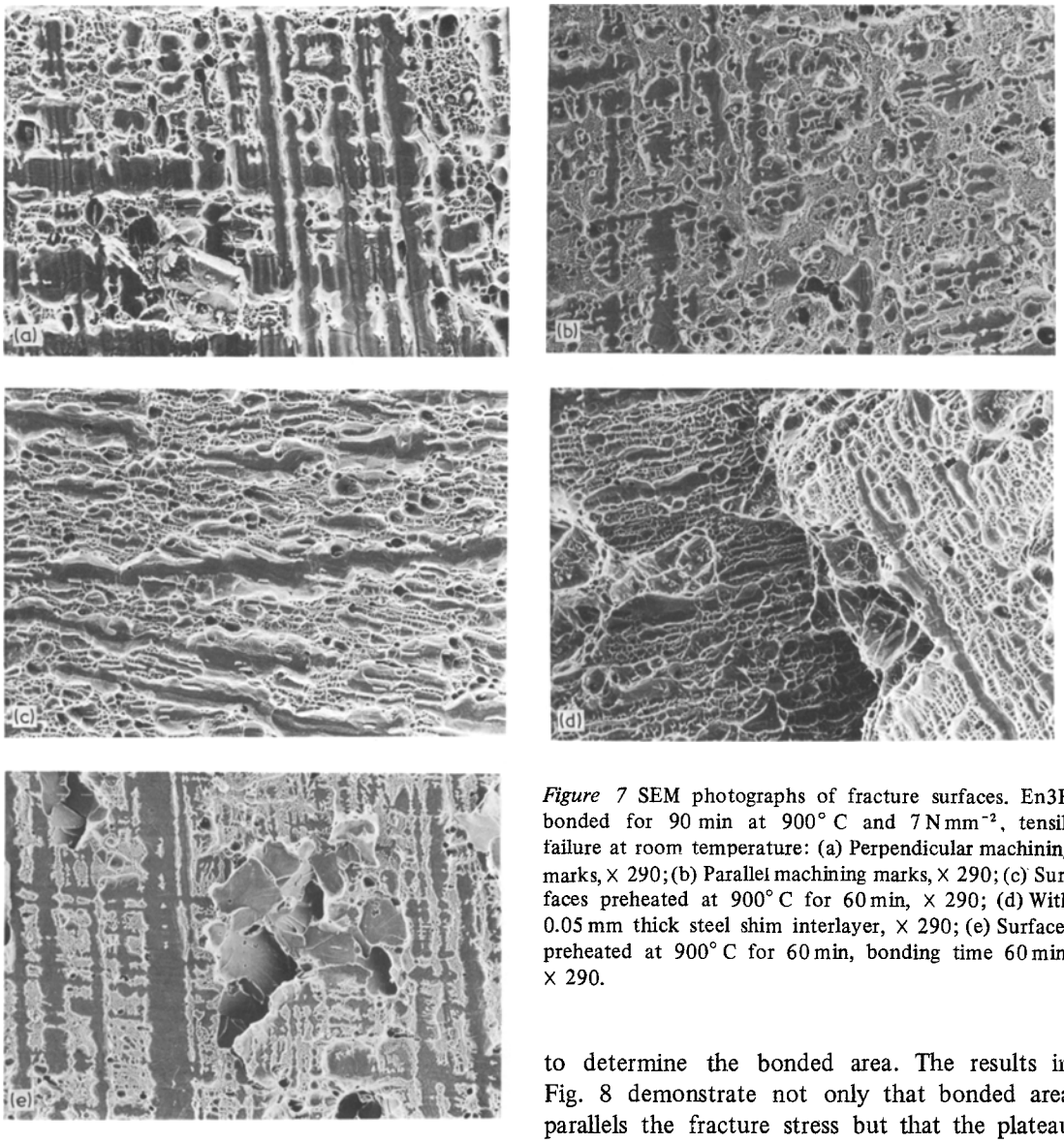



*Figure 7* SEM photographs of fracture surfaces. En3B bonded for 90 min at 900° C and 7 N mm$^{-2}$, tensile failure at room temperature: (a) Perpendicular machining marks, × 290; (b) Parallel machining marks, × 290; (c) Surfaces preheated at 900° C for 60 min, × 290; (d) With 0.05 mm thick steel shim interlayer, × 290; (e) Surfaces preheated at 900° C for 60 min, bonding time 60 min, × 290.

show that the effect on mechanical properties is not very marked in the material used in this study.

### 3.4. Quantitative assessment of the bonded areas

A measurement of bonded area might give an indication of the quality of a diffusion bond, thus bonded area assessment was carried out as described in Section 2.6. It was found that block tracing gave some correlation between percentage area bonded and the bond tensile strength, Fig. 8. Fractured En3B specimens from earlier trials [4], bonded at 1000° C and 7 N mm$^{-2}$, were examined to determine the bonded area. The results in Fig. 8 demonstrate not only that bonded area parallels the fracture stress but that the plateau in fracture stress is achieved at no more than 80% bonded area (the fracture stress plateau is close to the parent-metal strength). Specimens began to fail at some distance away from the bond after the longer times in these earlier trials [4]. As a bonded area greater than 80% has yet to be observed, even after lengthy bonding times at the higher temperature of 1000° C, it must be concluded that elastic constraint contributes significantly to the measured fracture stress as a consequence of the establishment of a triaxial stress state in the junctions between the voids.

In view of these results, a more sensitive property (than tensile strength) to be measured against bonded area would most probably be ductility, impact resistance or fracture toughness.

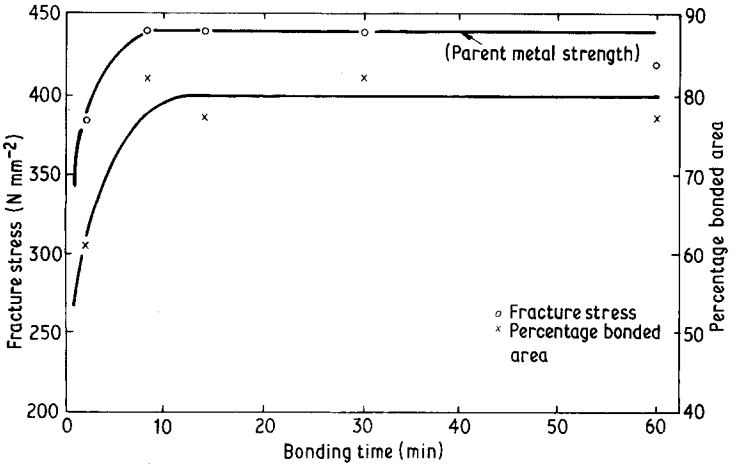



*Figure 8* Variation of fracture stress and percentage bonded area with time for En3B specimens from Stockham [4], bonded at 1000° C and 7 N mm$^{-2}$ pressure.

## 3.5. Metallographic examination

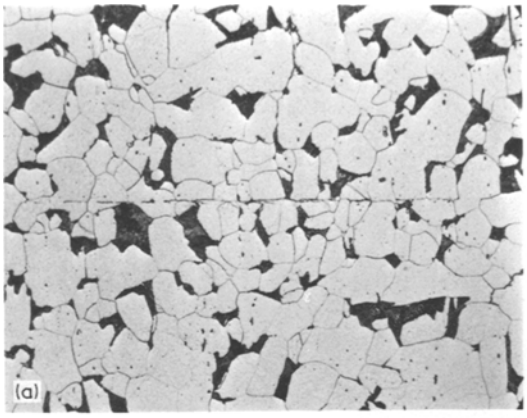
A number of points arise from the metallographic examination of bond cross-section (Figs 4 and 9).

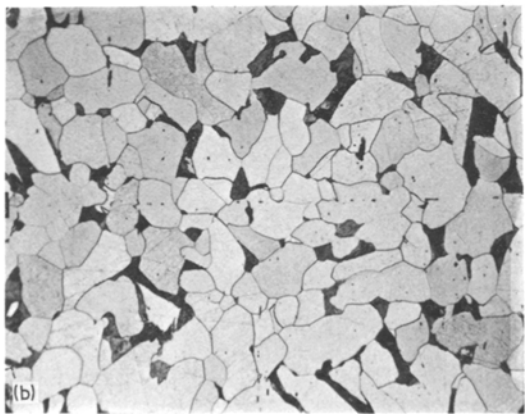
(i) The diffusion bonded material (i.e. that subjected to the thermal cycle) has an equiaxed grain structure which is finer than the as-received parent material.

(ii) After short bond times (2 min), the grains immediately adjacent to the bond interface appear to be smaller than those in the bulk (Fig. 9a), whereas, after longer times (120 min), grain growth has occurred (Fig. 9b) and no significant difference can be seen between the grains at or away from the interface. The refinement in grain size is restricted to a very narrow region next to the bond.

(iii) Non-bonded areas remain even where ferrite grains apparently obliterate the bond interface. Their appearance is that of voids remaining in the centre of the grains, as can be seen in Fig. 4, although it should be borne in mind that these apparent "voids" are probably sections through long grooves, as has been discussed earlier (Section 3.1.).

(iv) In one cross-section of an as-bonded specimen, a particularly sharp notch is seen, Fig. 9c, across which the grain boundaries appear to



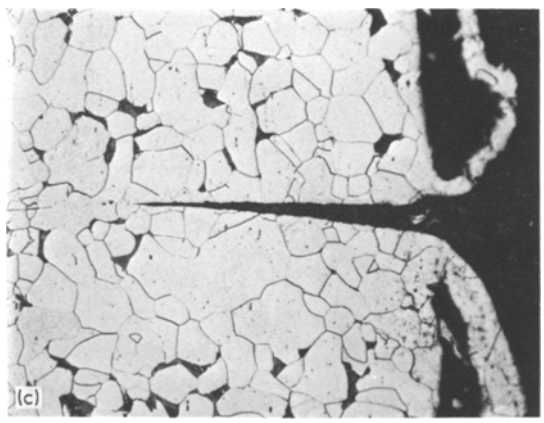




*Figure 9* Cross-section through En3B specimens bonded at 900° C and 7 N mm$^{-2}$: (a) Bonding time 2 min, × 203; (b) Bonding time 120 min, × 203; (c) Bonding time 20 min, surface preheated at 900° C for 60 min, × 203.

match. This indicates that the two surfaces must have been in intimate contact during bonding but that residual stresses were sufficient to open up the bond subsequently. The fact that the parting follows the bond line implies that the original interface (or some trace thereof) must have been preserved only to fail later. Thus, apparent continuity of grains across a bond line is not necessarily a sufficient criterion for joint integrity. Furthermore, premature removal of bonding pressure could affect the integrity of the final joint by allowing residual stresses to re-open the interface. The factors causing the occurrence of this type of notch may be associated with the "edge effect" as discussed in Section 3.1. (iii).

(v) Pearlite colonies are not seen to cross the bond interface in this material, confirming the observations of Taylor and Pollard [8]. They surmised that at bonding temperatures above $Ac_3$, the austenite grains tend to form a planar grain boundary along the interface with no significant growth across it but that, on cooling, ferrite nucleates preferentially at this boundary.

The above observations indicate that a planar grain boundary forms in the austenite along the bond line but is unable to migrate, i.e. that it is pinned by voids, oxides, and/or contaminants. This implies that traces of the original bond interface remain – a view that is consistent with the effect described in (iv) above and also with the observation that bonds often fail along their interfaces when mechanically tested. It follows that if the original austenite boundary is *not* pinned, it is unlikely that a planar layer of ferrite would form at the interface on cooling below $A_3$.

## 4. Discussion

### 4.1. Introduction

The long-term aim of the current research programme is to identify the dominant mechanisms in diffusion bonding, for any specified bonding condition, in order to establish a theoretical model for the process; such a model should eventually be capable of predicting the stages and the conditions required to attain high quality bonds. Several aspects that are relevant to an understanding of the formation of a diffusion bond have been revealed by this work and will be useful in formulating the model. In addition, a number of points have emerged which, though not directly concerned with mechanisms of bond formation, have a significance both when considering commercial applications of diffusion bonding and when continuing research into the mechanisms involved. Thus, in the discussion that follows, the implications of the results will be considered both theoretically (i.e. relevance to the model approach) and practically (i.e. relevance to the practice and applications of diffusion bonding).

### 4.2. Grinding mark orientation

The results show that, in the early stages of bond formation, the relative orientation of any directional surface finish will affect the geometry of the initial contacting areas and, therefore, of the voids that eventually must be closed in the later stages of bonding. It appears to be more realistic to treat these voids as long grooves (cylinders) or interconnected pores rather than as an array of discrete voids as has been assumed in earlier work [2]. This geometrical difference will affect not only the predicted rates of void closure, but probably also the relative contributions from the various competing material-transfer mechanisms; it is therefore of considerable theoretical importance when modelling the process. It is also of practical importance when considering the effect of relative orientation on fracture stress because it is possible that parallel alignment leads to a slightly larger area of surface contact, and so of bonding. Despite the scatter in results, the fracture stresses obtained from specimens bonded with parallel alignment of grinding marks were slightly higher than those from a perpendicular alignment. Thus, to maximize tensile strength, a parallel arrangement should be used in practice. Although the effect of grinding mark orientation could have been investigated further using the block tracing approach (Section 3.4.), it was not pursued because of a more important implication of the results from the fracture stress measurements which is now discussed.

### 4.3. Bonded areas and mechanical properties

It appears that at a comparatively late stage in the bonding process, i.e. when the tensile fracture stress has ceased to increase with increasing bonding time, there is still a relatively large area of unbonded metal (possibly as much as 20% of the whole area). This figure implies that even where parent-metal strengths are achieved, and even when tensile specimens neck down and break away from the joint line [4], there may well be

an appreciable area of unbonded material at the original interface. Although these unbonded regions do not appear to affect the tensile strength of the specimens, it is reasonable to suppose that toughness properties will be reduced by their presence. If so, one implication from this work is that tensile testing alone (particularly of as-bonded specimens) is misleading and does not provide a sufficient assessment of the quality of diffusion bonds; additional mechanical tests should be used. (Naturally, this will be of particular relevance in practice, e.g. when establishing procedures for fabricating diffusion-bonded components to be used in commercial applications). For instance, if impact strength is the controlling criterion, it is possible that more extreme bonding conditions than those required by tensile testing alone will be necessary for the attainment of good impact properties. This view is supported by Makara *et al.* [9] in which it is stated that the conditions they used for bonding high tensile steels do not necessarily produce good impact properties, although the latter can be achieved if bonding parameters are suitably chosen. Ushakova [10], working with stainless steel, reported that although tensile strength reaches a plateau with increasingly severe bonding conditions, the impact properties peak at an earlier stage. Thus, there appears to be more latitude in the achievement of good tensile rather than good impact properties.

It is, of course, possible that 100% bonding is *not* required to achieve good impact properties; if so, then the required degree of bonding will be an important parameter in any modelling work, and the hitherto accepted assumption, that complete void closure is mandatory [2], can be abandoned.

Another feature which might be expected to affect toughness is the unbonded region seen around the circumference of as-bonded specimens which had been fractured by impact at liquid nitrogen temperatures, Section 3.1. (iii). The unbonded region is believed to result from residual stresses originally introduced into the steel bar during its fabrication and then relieved during bonding; if so, a suitable heat treatment prior to bonding would considerably reduce these stresses and so eliminate the effect. The possible formation of this edge effect, and its ability to act as a stress raiser, again reinforces the need to use impact testing when assessing bond quality. Certainly, if diffusion bonding is to be considered for applications in which there is no post-bond machining (as often will be the case), the potential effects of such unbonded regions on mechanical properties must be taken into consideration.

Finally, there is a need for a non-destructive estimate (NDE) of the fractional area bonded. The case for NDE is strengthened by the observation that – at least in the one set of conditions that was examined – a direct correlation was found between bonded area and fracture stress.

### 4.4. Bond formation and possible embrittlement

Another postulate arising out of the work is the possibility that the joint zone is affected by the presence of oxide and contaminants originally at the surface. If so, several alternative modifications to bonding practice can be proposed: (a) the use of post-bond heat treatments to allow long-range dispersal of the contaminants by diffusion; (b) *in situ* surface cleaning immediately prior to bonding to remove the oxide and contaminants; (c) the use of coated surfaces.

### 4.5. Bonding and surface roughness

Other factors arising from surface phenomena during bonding stem from the effects of thermal smoothing. At this stage it is not possible to distinguish between, on the one hand, the possible beneficial effects of smoothing that should arise from an increased contact area, and, on the other, the possible detrimental effects that may result from a reduction of the initial pressure-welding of the fine asperities which will no longer be present. Furthermore, even if smoothing *per se* should prove to be beneficial, it may well be that the preheating required to produce it may, in fact, be detrimental. This could come about as a result of one or more of the following effects: (a) diffusion of surface contaminants; (b) a concomitant thickening of the ferrite layer at the interface; (c) the resultant decrease in stored mechanical energy in the surface.

While all these factors may be applicable at one stage or another in the bonding process, present knowledge is too limited to decide between the relative importance of each of them in relation to the final, desired properties of the joint.

Surface roughness measurements revealed a major undulation with a wavelength of some 100 $\mu$m, superimposed on which was a smaller wavelength of 15 $\mu$m itself carrying fine detail of about 1.5 $\mu$m wavelength. Although the effect of

surface wavelength on diffusion bonding has been discussed by Armong *et al.* [3], in fact they considered just two wavelengths, and then only in terms of their effect on void closure. In the present work, close examination of scanning electron micrographs of ground surfaces show that the fine (1.5 $\mu$m) detail consists of re-entrant ridges and laps. Although preheating removes these ridges and laps fairly rapidly, the fact that they are the only ones to disappear in a short time (and not the longer wavelength undulations) implies that there are no long-range effects. Thus, attempts to improve surface flatness by preheating ground surfaces are unlikely to succeed.

## 5. Conclusions

This work has demonstrated that the study of (fractured) interface surfaces of diffusion bonds can reveal considerable information which is of assistance both when attempting to derive a theoretical model of the process and also when selecting criteria with which to assess bonds for practical applications. It is evident that further work must be undertaken before a full understanding of even the simplest bonding case is achieved. However, the work reported here has shown that the following are of significance when diffusion bonding mild steel:

(i) that unbonded regions form as channels or grooves rather than as isolated voids;

(ii) that the relative orientation of these channels probably affects the mechanical strength of the bond;

(iii) that a large extent of unbonded area (of up to 20%) can occur in a weld with parent-metal tensile strength;

(iv) that a circumferential notch may be present in as-bonded specimens;

(v) that, from the preceding observations, there is a strong argument for more extensive bond assessment criteria than the simple, currently employed, tensile strength;

(vi) that surface diffusion will very rapidly remove the finer machining asperities;

(vii) that an interfacial layer of ferrite implies that the original austenite grain boundary formed along the bond-line was pinned by voids, contaminants and/or oxide;

(viii) that, from (vii), a "trace" of the bond interface is preserved and may deleteriously affect mechanical properties.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