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Solid-state Diffusion Bonding of Metals

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Abstract

Diffusion bonding is finding increased industrial applications in joining such diverse materials as superplastic metallic alloys, metal matrix composites and the new high critical temperature ceramic superconductors. In this paper, scientific aspects of the technique, rather than applications, are considered for the joining of metals; modelling the process and bond assessment by mechanical testing non-destructive evaluation and microscopy are discussed.

KEY WORDS: (Diffusion Bonding) (Modelling) (Mechanical Testing) (Non-destructive Testing) (Microscopy)

1. Introduction

Interest in solid-state joining goes back over 5000 years; a cup and chalice decorated by diffusion bonding have been dated at 3200 BC (Tylecote ref. 1). However, it is particularly in the last twenty years that there has been a renewed commitment to using diffusion bonding industrially and, equally important, to understanding the process and its limitations. Given the long history of the technique, a comprehensive review is beyond the scope of this paper, and reference can be made to the review by Owczarski and Paulonis (ref. 2) or to computer searches of the literature. Rather, a number of key areas of scientific interest will be considered based on the experiences of the author and his colleagues in the above department. Also, the use of interlayers or surface coatings will not be addressed despite their increased importance use, say for joining aluminium alloys, since the subject warrents its own separate review. There is, of course, an inherent danger in the approach taken in that attention will be focused on what is not well understood rather than on demonstrating how successfully the diffusion bonding process can be used in industrial applications! Accordingly, the final section of the paper lists a number of applications.

2. Process description

Diffusion bonding is defined as a process in which virtually no macro plastic deformation takes place, i.e. not more than a few percent of the components being joined. This distinguishes the process from cold pressure welding in which extensive plastic deformation (generally greater than 40%) is required. However, the definition is not rigorous and there is potential overlap between hot pressure welding and diffusion bonding, and also between brazing and variants of diffusion bonding in which liquid phases are formed, namely eutectic or transient-liquid phase (TLP) bonding. In TLP bonding, an interlayer or surface coating of different composition to the parts being joined is used so that after initial interdiffusion at the bonding temperature, the composition at the interface changes and a small amount of liquid phase is formed. This liquid phase ensures excellent surface contact between the parts being joined and can also assist in oxide removal. Further interdiffusion at the bonding temperature then causes the composition to alter further so that solidification occurs isothermally.

For solid-state diffusion bonding, joining is achieved mainly by atomic transport at the mating interfaces so that bond-line voids are eliminated. Thus the process variables of temperature and time must be selected to ensure high atomic mobility while pressure has to be restricted to values below which macro plastic deformation is prevented. In practice, this means that long periods of time (up to several hours) at moderate temperatures (one-half to two-thirds of the melting point) and low pressures (well below the room temperature yield stress) are often used, although bonding for short times can be successful for some materials. Surface contamination must be avoided and so bonding normally takes place under vacuum or an inert atmosphere. Clearly the nature of the material being joined has a strong influence on the selection of bonding conditions; precise chemical composition and grain size will affect the rates of atomic transport while the nature of the surface finish influences the scale of the initial bond-line voids that have to be eliminated.

For TLP bonding, the above process variables are still relevant. However, the selection of the surface coatings/interlayers and bonding temperature are, naturally, of over-riding importance in order to allow the formation and subsequent isothermal resolidification of the transient liquid phase. However, since TLP bonding relies on the presence of a liquid phase, it is not discussed further in this paper.

3. Modelling solid-state diffusion bonding

3.1 Aims of modelling

Two major aims in modelling diffusion bonding are to optimise the selection of the process variables for a given material and to provide an understanding of the mechanisms by which bonding is achieved. With regard to the latter, diffusion bonding takes place as a consequence of one or more competing mechanisms, the contributions from each changing not only from bond to bond as process variables and/or materials are changed but also during the time an individual bond is fabricated. Given this complexity, the existing models of diffusion bonding described in the following section tend to assume that the surfaces to be joined are free of contaminants and oxide, that bonding occurs between similar materials and that the materials are single-phase metals. These points are discussed further in the section on the limitations of existing models.

3.2 Models for single-phase metallic alloys

The majority of models simplify the geometries (profiles) of the contacting surfaces and assume that they can be regarded as a series of identical asperities touching at their tips to create an array of identical voids. The manner by which one (or part) of such a representative void is removed is then studied in the various models.

In 1966, Cline proposed that diffusion bonding could occur in two ways (ref. 3). In the first, the applied load was sufficient to cause considerable localised plastic deformation of the asperities. In the second, bond development was dominated by diffusion-controlled processes, more akin to the original recrystallisation model of Parks (ref. 4). The combination of these two approaches, albeit refined considerably, provides the basis for all subsequent models; most later quantitative models assume that localised plastic deformation occurs initially until the contacting areas of the deforming asperities increases sufficiently so that the stress on these areas falls below the yield stress of the material, and that subsequently bonding continues by various diffusional processes (including power law creep).

The next qualitative model was proposed by King and Owczarski (ref. 5) who suggested that bonding takes place in three stages. The first two stages were as above except that migration of the interface away from the voids was assumed to occur during the second stage so that volume diffusion was required in the third stage to ensure removal of the remaining isolated voids. In fact, interface migration does not necessarily occur in all materials although it has been observed in some instances, e.g. in the bonding of steel at temperatures well into the austenite region (ref. 6).

The first quantitative model of diffusion bonding was proposed by Hamilton (ref. 7), using the collapse of a series of long ridges which represented the contacting asperities. The model was based solely on plastic deformation and creep, and favourable agreement between experiment and prediction was found for the titanium alloy Ti-6 wt.% Al-4 wt.% V. However, in such an alloy and with the assumed surface roughness (large wavelength of 100 microns), it is evident (with the benefit of hindsight, obtained from subsequent modelling!) that creep would be expected to be the dominant mechanism under normal diffusion bonding conditions. Accordingly, any model based on just this one diffusion-controlled mechanism, namely creep, would be expected to give good results; the agreement might be poorer for other materials or if bonding conditions were chosen such that creep was no longer the dominant mechanism.

Garmong et al. (ref. 8) refined Hamilton's model by treating each ridge as a series of horizontal slices and then summing the response of each slice to the applied stress, and also by including diffusion mechanisms. In addition, the surface was modelled as two superimposed wavelengths, a feature that has not appeared in more recent models even though it more accurately represents a real surface in terms of both waviness (large wavelength) and roughness (short wavelength equivalent in later models to the dimensions of a ridge). The collapse of the two wavelengths was treated sequentially by plastic deformation and creep. The removal of the final small voids was modelled as the shrinkage of a thick-walled hollow plastic sphere under hydrostatic pressure; the various diffusion mechanisms were analysed using sintering equations (Coble, ref. 9). This model by Garmong et al. embodied not only the important features of earlier models but also stands out for several reasons:

- it includes the use of two wavelengths to describe surface roughness and this approach perhaps should be adopted again in future models given that long-range waviness may control the ease of bonding;
- it was the first to use sintering theory to model void closure, an approach that has been developed considerably in later models;
- it was potentially applicable to materials other than the titanium alloy for which experimental data were given.

The use of sintering theory was adopted more extensively in the model by Derby and Wallach (ref. 10, 11). One aim of their model was to be able to relate the extent of bonding for any single-phase metallic alloy to the various process variables and so examine the mechanisms by which bonding was achieved in different alloys. The original interface was regarded as a series of straight-sided ridges contacting peak-to-peak; this simplified modelling since plane strain conditions could be assumed in the subsequent analysis. Since diffusion bonding does not take place by one mechanism alone, seven possible mechanisms (illustrated for an elliptical void in Fig. 1) were considered.

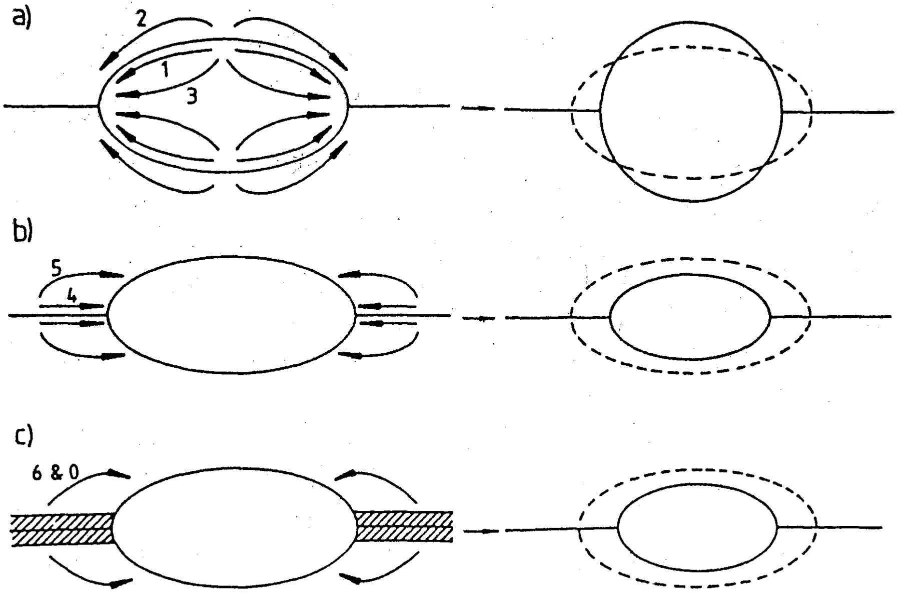


Figure 1. Routes of material transfer:
a) surface source mechanisms;
b) interface source mechanisms;
c) bulk deformation mechanisms.

The mechanisms were:

0. plastic yielding deforming an original contacting asperity;
1. surface diffusion from a surface source to a neck;
2. volume diffusion from a surface source to a neck;
3. evaporation from a surface source to condensation at a neck;
4. grain boundary diffusion from an interfacial source to a neck;
5. volume diffusion from an interfacial source to a neck;
6. power-law creep.

Given an initial geometry (i.e. wavelength and height of voids, based on measurement of the original surface roughness), rate equations (one for each of the seven mechanisms operating independently) are summed in the model to give an overall void shrinkage rate for a small increment of time and this is used to predict the small extent of bonding for the chosen time increment. This bonding naturally causes the void geometry to change by a small amount during each iteration (or small time increment) in a manner which depends on the extent of the contribution from each mechanism. Thus the bondline geometry can be revised appropriately and the extent of bonding for the next time increment is again predicted from the rate equations. This iterative process continues until bonding is complete. The results are displayed in a manner similar to that adopted by Ashby (ref. 12) and Wilkinson and Ashby (ref. 13) for pressure sintering, i.e. graphs of extent of bonding against one of the major process variables (temperature, pressure, void wavelength or void height) are plotted with time and/or rate contours superimposed on the plots (similar to Fig. 2, described below in more detail). In this model, three stages of void closure were employed;

i instantaneous localised plastic flow (as described at the start of this section);

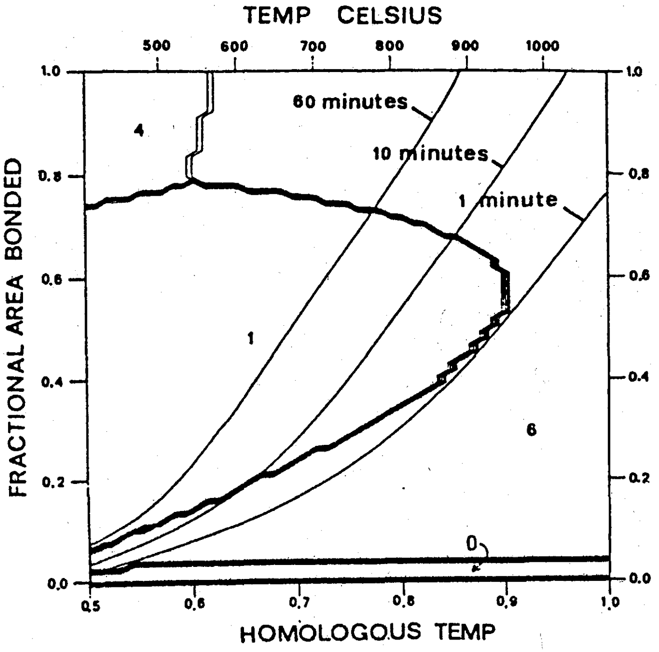
ii straight-sided void geometry until the void aspect ratio reached unity;

iii circular cross-section void (hence mechanisms for which the driving force arises from variations in surface curvature, e.g. surface diffusion, can no longer contribute to bonding).

Good agreement with experiment was observed for a number of materials (in particular, steel and copper), although the change in geometry from second to final stage resulted in a discontinuity in bonding rate. This discontinuity was reduced, but not eliminated, by remodelling the creep mechanism (Derby and Wallach, ref. 11). In commenting on an early version of this model, Allen and White (ref. 14) pointed out that the original assumption of long cylinders in the model by Derby and Wallach would lead to an overestimate of bonding time. In practice, a long cylinder might be expected to break up into a series of isolated spherical voids, the shrinkage rate of which would be considerably greater (by 5 to 20 times) than that of the corresponding cylinder. Additional points by Allen and White related to the effects of contaminants and, in a qualitative way, to grain size.

In 1984, Pilling et al. (ref. 15) proposed a diffusive creep cavitation model, based on original work by Chen and Argon (ref. 16) rather than using sintering as an analogue. An initial circular void geometry was adopted, instead of the elliptical shape generally observed in most micrographs of void cross-sections, since it was assumed that rates of the various surface diffusion mechanisms were sufficient to maintain circular voids. Thus the contributions from mechanisms whose driving force relies on differences in surface curvature are precluded and so the model is not as universal as those in which surface curvature mechanisms are considered. Since surface diffusion mechanisms are important when bonding many materials (e.g. steels under most conditions, copper at low temperatures), this model seems to be intended mainly for predicting bonding of fine-grained superplastic alloys in which creep mechanism tend to dominate. Nonetheless, an innovation of the model by Pilling et al. was the inclusion of an analysis of grain size; the material surrounding each void is divided into horizontal slices, each of thickness equal to the grain size, and the contribution of each boundary intersecting the void added to the grain boundary contribution.

The final model to be described was developed in 1984 by Hill (ref. 17) and is to be published shortly (Hill and Wallach, ref. 18). It aimed to overcome some of the approximations and limitations inherent in the previous models. Firstly, by using an ellipse to represent the initial void shape, the need for discrete stages with their own geometries is removed since an elliptical cross-section void can develop into a round void by successive incremental changes to the two axes of the ellipse, i.e. without a discontinuity. In fact, a further advantage of describing a void as an ellipse is that surface source mechanisms can be re-activated naturally if a void which has become circular during bonding then changes its shape back to an ellipse as a consequence of contributions to bonding from interface sources and creep mechanisms. Secondly, the elliptical geometry permits more rigorous analyses of plastic deformation and creep than those utilised earlier by Derby and Wallach (ref. 10,11). Thirdly, a simple statistical analysis enable the contributions from grain boundary sources in fine-grained materials to be more accurately included and so the effects of grain size can be incorporated. In other ways, the same philosophy as proposed by Derby and Wallach was continued; the contributions from the seven possible bonding mechanisms (shown in Fig. 1) were summed iteratively and the results displayed in a similar fashion by a mapping technique or more simply by plotting fractional bonded area against time. The mapping approach is shown in Fig. 2; the axes are fractional bonded area against one of the five process variables (temperature, pressure, surface roughness height, surface roughness aspect ratio or grain size). Time contours



Pressure = 5.6 MPa
Grain size = 100 µm
Surface roughness
Height = 3 µm
Wavelength = 50 µm

Figure 2. Output from mainframe computer model for diffusion bonding copper under the process conditions shown. The number in each area denotes the dominant bonding mechanism as above.

and/or rate of bonding contours are superimposed on the maps. When one process variable is selected for the x-axis, the remaining four are set to specific values which are input with the other material data. The resulting output is a diffusion bonding map in which discrete areas show the dominant mechanism; a number (from 0 to 6, corresponding to the seven mechanisms listed above) in each area identifies the dominant bonding mechanism. The simpler graphical output in which bonded area is plotted against time (see Fig. 3) has the advantage that the relative rates of the various competing mechanisms can be clearly seen; with the mapping approach, the dominant mechanism only is shown and, in an extreme and hypothetical case, the dominant mechanism might be contributing only marginally more than one seventh of the total bonding rate. Nonetheless, the advantage of the mapping approach is that the effect of altering a process variable (that chosen for the x-axis) is immediately seen from just one map whereas in the simpler graph all five process variables have to be specified. Good agreement with experimental data is reported (Hill and Wallach ref. 18).

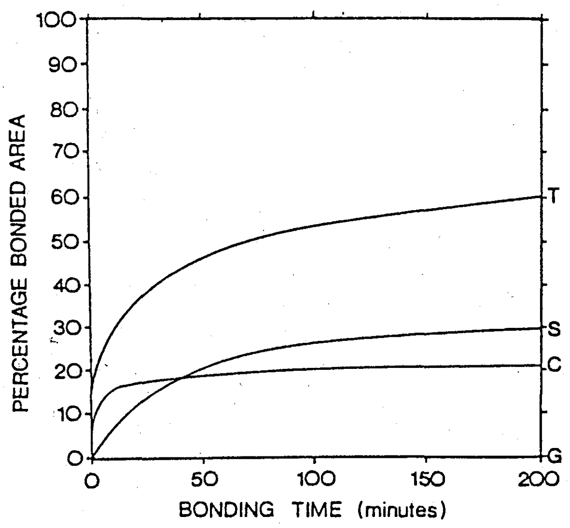


Figure 3. Output from the microcomputer model for diffusion bonding copper under identical conditions to those in Fig. 2 and the temperature of 900°C. The contributions to bonding are identified by T (total amount of bonding), S (surface sources), G (interface sources) and C (creep and plastic deformation).

3.3 Current status of modelling

The recent quantitative models of diffusion bonding claim good agreement with experimental data. Thus, for single phase metallic alloys, it would seem that there now is a reasonable understanding of the dominant bonding mechanisms for different materials and bonding conditions, and of the extent to which each mechanism contributes towards bonding. The use of a model can, therefore, lead to optimisation of process variables in a much faster manner than can be achieved by experiment (fabrication and testing of actual bonds), with the proviso that bonding is still needed to confirm the results of any model predictions!

In fact, it is now worth asking whether additional refinements to such models can lead to marked improvements in agreement with experiment. Firstly, there is the problem of evaluating the predictions of any model, i.e. the experimental difficulty of accurately measuring bonded area (for a summary, see Derby, Briggs and Wallach ref. 19). Secondly, there are difficulties in obtaining, with sufficient accuracy, the precise input data for modelling a specific material, e.g. diffusion and creep data generally are obtained from tabulated data and not measured exactly for the material being bonded (the thermo-mechanical history and precise surface condition will differ for each nominally identical batch of the same alloy). With only slight adjustments to process parameters and input data, model predictions can change considerably from poor to exact agreement with the results obtained from bonding experiments. This has been considered in detail by Hill (ref. 17) who has shown that for many materials the values of shear modulus, creep and surface diffusion data (and their temperature dependence) are critical. The latter two values are particularly important since, for most materials, creep and/or surface diffusion mechanisms dominate and contribute significantly to bonding. Unfortunately, it is precisely these two properties that are difficult to measure accurately and which can vary from batch to batch of nominally identical material.

Despite the above reservations, modelling of diffusion bonding in the last decade has considerably improved understanding of the mechanisms involved. This knowledge not only helps to identify bonding conditions, so minimising the experiments that have to be carried out, but also can suggest different conditions under which a particular alloy might be bonded to the same extent. In addition, in some cases, alloy composition can be adjusted so as to optimise or enhance the rate of the dominant bonding mechanism, and so promote bonding. As fabrication routes become more critical for highly alloyed materials, modelling can help the process itself to be more widely used.

3.4 Future of modelling

The optimism of the concluding paragraph above must be tempered by the awareness of what still needs to be considered in further work on modelling diffusion bonding. To date, most models have been limited to single-phase metallic alloys and have treated the surface in a very simple manner. For instance, the diffusion bonding of aluminium alloys is of considerable commercial interest and, if the surface oxides are ignored, the various models predict that bonding would be easily achieved under conditions that would be accepted by industry. In practice, the tenacious surface oxides on aluminium alloys inhibit bonding and also result in unacceptable mechanical properties; bondline failures are often observed with low strength values. Thus the challenges of the future include:

- improved surface profile descriptions (e.g. returning to the two wavelength model of Garmong et al., ref. 8);
- recognition of the roles of surface contaminants and surface oxides (which may necessitate considerable experiment just to understand what occurs at the initially contacting surfaces);
- modelling of the two-phase materials (improving on the use of the simple rule of mixtures, i.e. averaging the properties of the various phases according to the amounts present in an alloy);
- allowing for the effects of thin (less than a micron thick) surface layers as the use of such layers (by sputtering) becomes more prevalent e.g. when bonding materials with insoluble surface oxides such as aluminium alloys;
- modelling of bonding between dissimilar materials;
- modelling of transient liquid phase bonding;
- modelling of bonding in non-metals, and between metals and non-metals.

Given the advances of the last decade and the renewed interest in diffusion bonding, progress in at least some of the above is anticipated!

4. Mechanical testing of diffusion bonds

Mechanical testing can be used to determine bond properties both to see whether a joint is suitable for a particular application and also to measure the extent of bonding (thus providing a means of assessing the predictions from diffusion bonding models). When fabricating diffusion bonds, one aim is to produce bondline microstructures which are identical to those of the parent metal since the mechanical properties might then be expected to be identical to those of the parent metal. This is a potential advantage of the process, in contrast to fusion welding processes in which the solidified welds have quite different microstructures and hence different mechanical properties from parent metal. In certain metals (e.g. titanium, steel and copper), it is possible to produce diffusion bonds with microstructures apparently identical to those of parent metal. However, while some mechanical properties then are found to be similar to those of the parent metal, this is not necessarily the case. Hence it is extremely important to consider various test methods and test specimen designs. The selection of a test method will depend on the information being sought, the material being tested and the ease of carrying out the test. For instance, failure at the bondline is required if bond strength is being measured whereas failure away from the bondline might be regarded as an appropriate indication of success if a fabricated component is being evaluated. Various mechanical tests have been used to evaluate the strengths of diffusion bonds and the following is intended only to be representative of the methods reported in the literature.

Tensile tests are relatively simple and parent metal tensile properties have been reported for diffusion bonds, for instance, in titanium alloys (ref. 20) and in steels (ref. 6). However, parent metal tensile properties can be obtained even when up to 20% of the interface remains unbonded (ref. 21,22) and so parent metal failure does not necessarily reflect complete bonding.

For bonds between thin sheets, conventional tensile tests are generally impracticable although Pilling and Ridley (ref. 23) have developed an annular tensile test for sheet specimens; their design was a variant of the slotted test piece commonly used in shear tests. In general, the strength of bonds in thin sheets can be evaluated by shear tests and the simplest specimen geometry is that of a single overlap. However, as this geometry results in out of plane bending and the generation of peel stresses during testing, various improvements to the method have been suggested. For example, clamping of the specimen to rigid supports (ref. 24) and the use of teflon-coated clamping plates (ref. 23) have been reported. Slotted and double overlap test specimens of various designs can be used although the machining is both expensive and critical. Moreover, stress concentrations at the roots of the machined slots can considerably influence the measured strengths. An additional disadvantage is that the area under test is relatively small since the overlap length must be less than about 80% of the thickness of the sheet being joined if failure along the bond line is to occur.

The use of peel tests for thin sheet (e.g. ref. 25) has not been widespread despite their extensive usage to test adhesives (ref. 26) and yet this test potentially offers advantages. Rather than testing a small area (as in overlap specimens), peel tests can provide mechanical strength data from the entire bonded area; as the test progresses, strength variations across an individual bond will be apparent while the average result will be more representative of overall bond quality. In addition, there is considerable interest in using diffusion bonding combined with superplastic forming for joining aluminium alloys (e.g. see ref. 23) and especially the new Al-Li alloys. The use of a peel test carried out at elevated temperatures potentially provides a more representative assessment of whether a bond can withstand a subsequent superplastic forming operation. Thus, it is anticipated that peel tests will be used more extensively especially as research into the bonding of aluminium alloys continues.

One of the consequences of diffusion bonding is that a planar array of small voids can be left at the bond line. The fact that tensile failure may still occur away from the bond line when the bonded area exceeds 80% and such voids still exist can be misleading; the implication is that complete bonding has occurred. While this result can be explained in terms of the plastic constraint at the bond line arising from the presence of the voids, other mechanical properties may well be adversely affected. Accordingly, bond assessment by methods which will be sensitive to the presence of the voids, e.g. fatigue and/or impact testing, must be considered to ensure complete bond integrity. The use of fatigue testing has been reported for a number of systems including nickel (ref. 27) and titanium alloys (refs. 28,29). Impact testing has the advantage that it is a relatively quick test method and its use, particularly for steels, has been extensive. Typical results for plain carbon steels have been published by Thornton and Wallach (ref. 6,30) and Charpy impact strengths of the bonds reached 60-70% of the parent metal values. It must be remembered that such values would be regarded as good for a fusion weld but, given that diffusion bonding is theoretically capable of parent metal microstructures (hence properties), an explanation is required. To account for these results for plain carbon steels, it has been proposed (ref. 30) that two competing events occur as bonding proceeds.. The first is that the voids which form as the two original surfaces come together shrink as a consequence of the mass transfer mechanisms discussed in Section 3. At the same time, and in competition, impurity elements such as S, Sn, Sb, P diffuse to the void surfaces (which form preferential sites for such elements) and stabilise the voids. As the voids become smaller, the latter mechanism becomes energetically more favourable and so the voids are prevented from shrinking entirely; the resulting array of bondline voids (with diameters of nm magnitude) accounts for the lower impact properties despite the apparent parent metal microstructure.

With regard to using mechanical tests to assess the accuracy of predictions from diffusion bonding models, there are two problems. The first is the fact mentioned above, namely that parent metal tensile strengths can be achieved even when only 80% of bonding has been achieved. Thus it is not possible to assess accurately the final stages of bonding. The second problem is that there often is scatter in the mechanical test results for nominally identical bonds. While the use of, say, a peel test might result in better averaging of measured properties, such a test is not always applicable. A more analytical approach, for instance the use of Weibull statistics (e.g. see ref. 31) is required. While this approach has been reported (ref. 23), a disadvantage is that it still is necessary to test a large number of specimens.

In summary, it is evident that a large number of potential test methods exist. However, the interpretation of results can be misleading and, in general, more than one type of test method should be employed. Moreover, the tests should, ideally, reflect the intended service conditions for a bonded component.

4.3 Non-destructive test (NDT) methods

NDT methods are established for assessing fusion welds; ultrasonic testing, X- and gamma-radiography find widespread usage. However, these methods are of limited value for assessing diffusion bonds because the scale of the interface voids is too small. For instance, two studies (ref. 32,33) investigated the effectiveness of both X-ray and ultrasonic techniques and concluded that interfacial voids could not be observed unless they were of unrealistically large sizes. With regard to ultrasonics, the frequency used generally is in the range of tens of MHz. The speed of sound in metals is typically about 5000 m s^{-1}, which gives a wavelength in the range of about 100-200 μm for the ultrasound. As this wavelength is considerably larger than the defects expected in the latter stages of bonding, the defects cannot be detected. To overcome this limitation, the use of acoustic microscopy has been investigated briefly (ref. 19), since higher frequencies and focused beams can be employed. Although the technique is still in its development stage and, particularly in transmission mode, suffers from resolution problems due to scattering effects from microstructural features (e.g. grain boundaries, dislocation arrays), preliminary results were encouraging (ref. 19).

The electrical conductivity of a piece of metal is proportional to the cross-sectional area through which the current passes and so bond resistance should provide an inverse measure of bonded area. This approach has been evaluated by numerous workers (e.g. ref. 19, 34). While some promise is shown (ref. 34), the technique suffers from similar problems to the above NDT methods. Firstly, the variation in resistance as bonding proceeds is not a linear function (ref.19); while the initial stages of bonding can be easily monitored, say with potential drop equipment, the instrument sensitivity needed when bonding is greater than about 70% increases indefinitely. Extremely sensitive and accurate equipment is needed both to measure the potential difference across a bondline and also to ensure a sufficiently stable current through the test piece. It is possible that a system based on differential amplification would improve accuracy, i.e. comparing the signal across parent metal with that across the bondline. Secondly, variations in the parent metal microstructure can cause resistance changes of magnitudes similar to those expected from the presence of the very small voids at the bondline in the final stages of bonding; this makes interpretation of any signal extremely difficult. Nonetheless, the technique has found applications, for instance in the interpretation of the nature of the bonded areas at the bondline (ref. 34).

Thus it can be seen that various NDT techniques can reveal the presence of large defects or large unbonded areas. However, conventional methods (radiography and ultrasonics) are not sufficiently sensitive to detect the fine scale voids present in the latter stages of bonding. Unfortunately, it is precisely these fine scale voids that can be detrimental to mechanical properties such as toughness and so it is imperative that their presence is both established and, ideally, quantified. Other NDT techniques that show some promise (acoustic microscopy and electrical resistivity measurements) are not truly non-destructive in that the test samples have to be machined from bulk samples and their dimensions must be carefully controlled. Thus true NDT of diffusion bonds requires further development in order for it to provide a means to reliably assess bonds as they are are fabricated; this is an essential requirement for certain industrial applications.

4.4 Microscopy of bondlines

The microstructures of diffusion bonds have been examined at various levels of resolution ranging from light microscopy to transmission electron microscopy (TEM). Clearly, the choice of a particular technique reflects the nature of the information being sought as well as the ease of sample preparation (particularly for TEM). Morphological features such as interfacial voids, local grain structure and surface layers/oxides can all be identified depending on the scale of the information and the available magnification of the technique used. Not only can the bondline morphology be revealed but it also is possible, using approaches such as electron energy loss spectrometry (EELS) or X-ray microanalysis, to investigate composition changes across bondlines. More sophisticated surface analysis techniques also have been used to examine bondline chemistry; the use of Auger electron spectroscopy and laser microprobe mass spectrometry have been reported (ref. 30, 35, 36). These methods of investigation are required in order to understand the role of coatings and interlayers used to assist bonding materials such as aluminium alloys or oxide-dispersion-strengthed nickel-base alloys. For instance, it is only by studying, using TEM (with associated EELS and X-ray analysis detectors), the morphologies and nature of the oxide types at the bondlines of diffusion bonds in aluminium alloys (see Fig. 4) that a full understanding of the observed mechanical properties can be attained. Such an appraisal is also essential if potential methods for oxide elimination or removal are to be developed and fully evaluated.

5. Applications

The aim of this paper has been to consider the current state of diffusion bonding metals in the solid state, both considering the role that modelling can play in providing some understanding of the mechanisms by which bonding occurs and also considering the relative advantages and limitations of the available methods by which bonds can be examined. However, science sometimes lags behind technology in that a technique can find industrial application before a full and detailed scientific understanding is reached. The process of diffusion bonding, has, as stated in the introduction, been used for 3000 years and so is now widely accepted industrially. Among the applications (see also ref. 2) are:

aerospace components in both titanium and nickel-base alloys, either by straight diffusion bonding or coupled with superplastic forming;

industrial components in which differential thermal expansion tooling is used to provide the bonding pressure, e.g. piston crown assembly;

fabrication of metal-matrix composite structures, e.g. Al-B and Ti-B;

joining in order to minimise microstructural damage, e.g. directionally-solidified eutectic turbine blades or oxide dispersion strengthened alloys;

joining of dissimilar metals with or without the use of interlayers;

joining of large areas in one operation, e.g. copper heat exchangers;

fabrication where no macro deformation or post-bond machining is permitted;

manufacture of parts with sophisticated designs, e.g. with internal cooling channels.

6. Acknowledgement

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See next page.

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