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Journal of Nuclear Materials 283–287 (2000) 982–986

journal of nuclear materials

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Evaluation of the deformation fields and bond integrity of Cu/SS joints

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Abstract

The stress states that lead to failure of joints between Glidcop™ CuAl25 and 316L SS were examined using finite element modeling techniques to explain experimental observations of behavior of those joints. The joints were formed by hot isostatic pressing (HIP) and bend bar specimens were fabricated with the joint inclined 45° to the major axis of the specimen. The lower surface of the bend bar was notched in order to help induce a precrack for subsequent loading in bending. The precrack was intended to localize a high stress concentration in close proximity to the interface so that its behavior could be examined without complicating factors from the bulk materials and the specimen configuration. Preparatory work to grow acceptable precracks caused the specimen to fail prematurely while the precrack was still progressing into the specimen toward the interface. The finite element model demonstrated maximum stress concentrations in the interface layer to be shifted off-center. An additional benefit from the finite element modeling effort was in understanding if the stress states in this non-conventional specimen were representative of those that might be experienced in practice.

1. Introduction

Specimens of Glidcop™ CuAl25 and 316L stainless steel (SS) were taken from hot isostatically pressed (HIP) plates for testing in the unirradiated and the irradiated states. The HIP conditions were 982°C at 101 MPa for 2 h. The material compositions and the results of mechanical testing and microstructural analysis of this and other HIP bonded plates in the unirradiated conditions were reported previously [1]. For the purpose of developing a miniature specimen configuration for irradiation testing that would provide a reasonably severe loading condition for the joint, bend bar specimens with the joint inclined 45° to the major axis were fabricated. The specimen configuration and dimensions are shown in Fig. 1.

The manufactured specimen design was selected for its ability to provide a complex state of stress at the material boundary layers for experimental testing. The local stress-state was to be intensified by the presence of a crack tip very close to the interface. This necessitated the growth of a precrack from the starter notch to a length nearly half way through the specimen. This was to be accomplished by fatigue loading at moderate load levels. Specimens with such precracks were to be tested to failure by loading in three-point bending in both the unirradiated and the irradiated states. The bending loads were to be applied so that the lower side was supported at each end, and the upper side was loaded just opposite the starter notch in the middle of the upper surface. Specimens with a sufficiently long precrack were to be loaded monotonically to note their load-deflection response and to provide information about the failure mode (e.g. delamination, plastic bending or brittle failure through the base materials).

When attempts were made to initiate and grow the precracks, it was found that specimens either bent prematurely when the CuAl25 side was down (i.e. on the right side of the specimen shown in Fig. 1) or specimens delaminated prematurely when the SS side was down. This frustrated the development of a useful miniature

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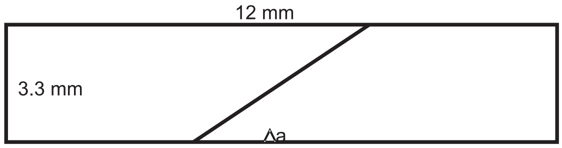

Fig. 1. Bend bar configuration showing the 45° inclination of the HIP joint to the major axis of the specimen and the position of the starter notch at the center of the lower surface of the specimen.

specimen configuration since it was not possible to grow precracks of appropriate length. It also brought the strength of the joint into question since it was not anticipated that the joint would fail in such a manner, particularly in the unirradiated, as-fabricated state. This led to a further analysis of the stress states induced by loading of this specimen configuration. These results were on unirradiated specimens; results on irradiated specimens of this configuration are presented elsewhere in these proceedings [2].

2. Modeling approach

To better understand the nature of the stress states and the failure modes, finite element analysis was performed on the specimen configuration shown in Fig. 1. Most of the analysis was done with the commercial finite element method (FEM) software packages ANSYS. The model employed 6-node triangular elements with a quadratic displacement behavior suited to model irregular meshes, which allowed larger element densities around the crack tip and concentrated stress/strain areas. Loads were applied in three-point bending. For consistency, results were compared for an applied load of 30 MN. Calculations were carried out for a number of precrack lengths, from 0.3 to 1.3 mm. The effect of material orientation was also considered such that calculations were performed for bend specimens with either the CuAl25 on the upper side and 316L SS on the lower side, or vice versa. In addition, a thin layer of pure copper between the two other alloys was considered to better reproduce the features observed at the HIP interface (see Ref. [1]).

The material properties for CuAl25, 316L SS and Cu were taken from those developed to model the initial plate materials, where good agreement was found between experimental results and other FEM modeling efforts [1]. The tensile behavior was modeled using a bilinear hardening model where the elastic loading portion was represented by the Young's modulus. The initial plastic, strain-hardening behavior was modeled with a yield strength and a plastic modulus. The values for the three respective materials are shown in Table 1.

3. Results and discussion

The results of the FEM analysis for several loading conditions and crack lengths were compared. For the purposes of discussion, the von Mises stresses for CuAl25 on 316L SS are shown in Fig. 2 for notch lengths of 0.65, 1.0 and 1.3 mm, respectively. The shear stress distributions are shown in Fig. 3 for notch lengths of 0.65, 1.0 and 1.3 mm, respectively, from the same calculations. The stress ratio, defined as the ratio of von Mises stress to yield stress is shown in Figs. 4 and 5 for CuAl25 to 316L SS and 316L SS to CuAl25, respectively. All plots are for a load of 30 MN. Several points of comparison are noteworthy in these plots.

The first major point of comparison is the relative distribution of stresses around the notch tip with the notch either in the CuAl25 compared to the notch in 316L SS. The broader stress distribution in the CuAl25 indicates that the strain-hardening behavior of the 316L SS is more severe than in CuAl25 despite the high-yield strength of the latter material used for modeling. This results in a more localized crack-tip stress field for the 316L SS compared to the CuAl25 and is consistent with the experimental observation that under fatigue loading a starter crack could be initiated in the 316L SS when it was on the lower, notched side. Under similar loading conditions, the bend specimens merely deformed without the initiation of a starter crack when the CuAl25 side was down.

A broad comparison of the stress states shown in Figs. 2 and 3 does not indicate the initial concept that the major stress concentrations would occur in the vicinity of the interface at the center of the specimen. In fact, the major stress concentrations are off-center, but

Table 1
Material properties for CuAl25, 316L SS and Cu used in the FEM analysis

Material	Elastic modulus (GPa)	Yield strength (MPa)	Plastic modulus (GPa)	Density (kg/m^3)	Poisson's ratio
316L SS	200	207	8.0	7860	0.283
CuAl25	130	400	11.0	8940	0.343
Cu	130	60	11.0	8940	0.343

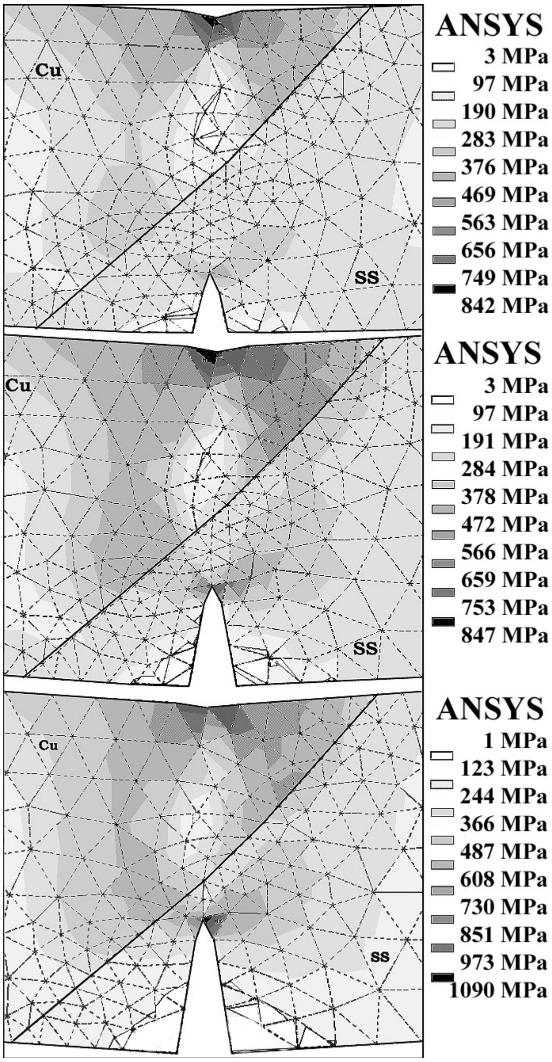

Fig. 2. Comparison of von Mises stress states as a function of crack length with various crack lengths. Note that Cu refers to CuAl25 and SS refers to 316L SS.

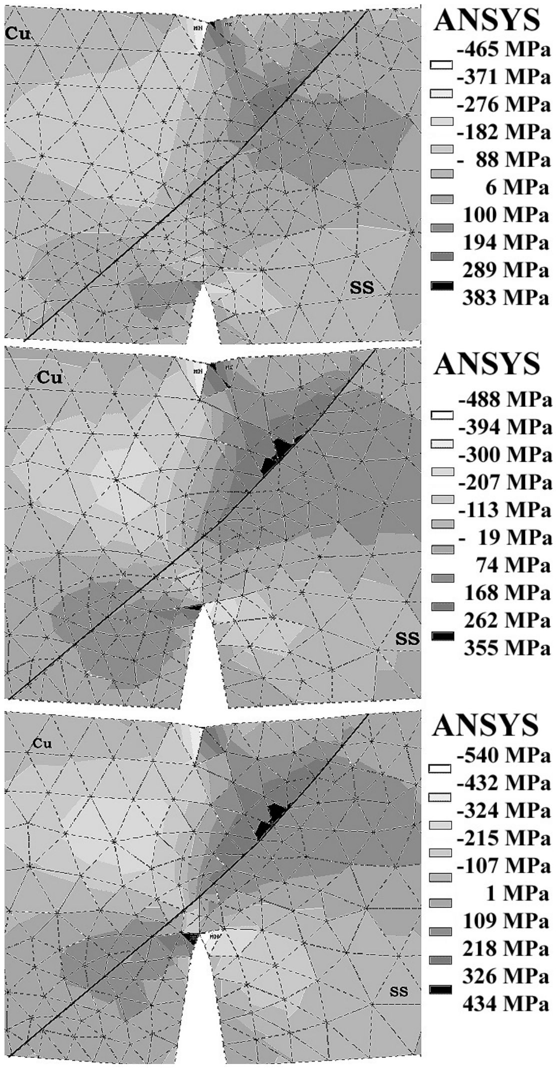

Fig. 3. Comparison of shear stress states as a function of crack length with various crack lengths. Note that Cu refers to CuAl25 and SS refers to 316L SS.

do develop along the interface as well as at the notch tip and the load points. The shear stress values are high along the interface for both specimen orientations, and are likely to be the major cause of the delamination as observed during precracking with the 316L SS on the bottom. The precise nature of the initiating conditions for failure by delamination is not known. Lap shear measurements on these joints indicated that they should posses a shear strength in excess of 100 MPa [1]; the calculated values here are in excess of that level. Nevertheless, some local or microscopic process for initiating a crack along the interface is required. A mechanism for the development of a starter crack has not yet been identified. It is also useful to note that the calculated stress concentrations at the points of intersection of the interface with the upper and lower surfaces are not especially large. From the calculations, it would appear unlikely that a crack would first initiate at the intersection with the surface. A second point to note is that the highest resolved stresses along the interface are in shear. Studies of interfaces in brittle materials indicate that tensile stresses may dominate the failure process [3]. The stress states leading to delamination in ductile materials is still not well clarified.

The deformation behavior of the CuAl25, as modeled here, is also important in assessing the potential modes of failure. Owing to its lower plastic modulus, it tends to deform over larger specimen volumes than does the 316L

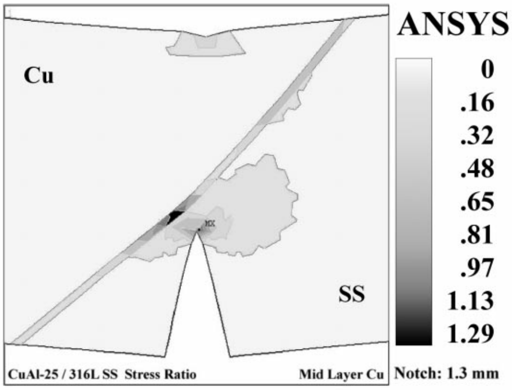

Fig. 4. The stress ratio in a CuAl25 to 316L SS HIP bonded plate with a layer of pure copper at the interface.

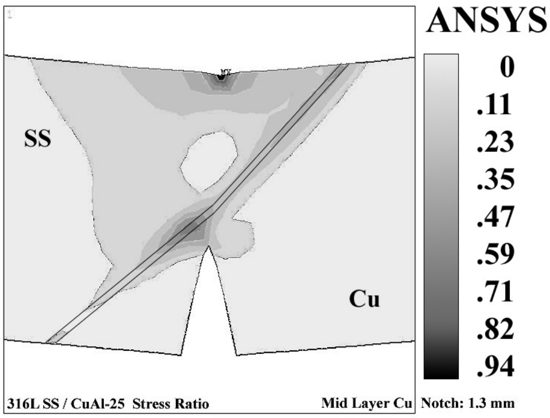

Fig. 5. The stress ratio for a 1.3 mm notch in a 316L SS to CuAl25 HIP bonded plate with a layer of pure copper at the interface.

SS. This tends to cause the shorter wedge of CuAl25 in the central section of the specimen to deform readily in comparison to the 316L SS. This leads to the formation of a plastic hinge through the CuAl25 resulting in an overall deformation of the specimen. While the conditions for delamination are likely to be useful as a guide for limiting stresses for ITER design considerations, the likelihood of plastic hinges forming through thin sections of CuAl25 is small in practice. Thus the information's usefulness regarding the bending of the specimen during precracking with the CuAl25 on the lower, notched side is of questionable value for design purposes.

The current study was limited to modeling behavior of specimens at room temperature, taken to be the stress-free state. In actuality, residual stresses due to fabrication or due to exposure at other temperatures should be considered. Mismatches in thermal expansion coefficients between the two materials would result in interfacial stresses that are not considered here. The magnitudes of residual stresses are likely to be much smaller than those imposed by the notch and loading conditions considered in these FEM calculations. In fact, the intended use of a sharp crack tip to concentrate the stresses at the interface in the current specimen design was to elevate local stresses at the interface. Available elastic analysis studies of a crack tip near an interface between dissimilar materials indicates that the dissimilar properties can either accelerate or retard crack propagation based on the relative values of the Young's moduli [4]. In the current case, it was not possible to bring the crack tip near enough to the interface to examine this possibility.

Further examination of the stress states and failure process in this specimen configuration should also account for the unique material properties of the interfacial zone. It was found in this and other experimental work [1] that the interfacial fracture takes place next to the interface in the diffusion zone of the CuAl25. As the inter-diffusion zone has a microstructure that differs from either base material, it is appropriate to consider it as a separate material. Further modeling shows the effect of the interface. In Figs. 4 and 5, the addition of a thin layer of Cu between the two alloys is considered. It should be noted that the effect of the copper layer, with its low-yield properties, further concentrates deformation in the interfacial layer. Typically, yield (or other non-reversible deformation process) will occur first in the interfacial layer.

4. Conclusions

Finite element analysis was performed on a miniature bend bar specimen comprised of a bilayer of CuAl25 and 316L SS; in several cases, the influence of an interfacial layer of pure copper was also considered. The specimen was designed such that the interface was inclined at 45° through the specimen. The lower section contained a notch or precrack that was intended to concentrate a load on the portion of the interface at the center of the specimen in order to test its mechanical properties. The FEM analysis indicated that the copper alloy tended to deform more readily than the stainless steel. This results in the spreading of the plastic strain over a larger portion of the specimen when the copper alloy is on the bottom. This leads to the formation of a plastic hinge and the specimen fails by excess ductile deformation, an unlikely event in component applications. When the stainless steel is placed on the bottom during loading, the stresses are concentrated more locally, and this permits the development of a crack. In both cases, however, the maximum stress concentrations on the interface are not at the center of the specimen, rather they are shifted off-center. In addition, the major stress concentrations appear to be dominated by the shear components and this likely supports the initiation of delamination, as noted in precracking specimens with 316L SS on the bottom. The presence of a thin interfacial layer of pure copper, consistent with microstructural observations of the real joints, leads to local yielding at the interface prior to general yielding in the bulk materials. This further exacerbates the likelihood of failure at the interface. The precise mechanism for delamination of the interface cannot be resolved from the current analysis, but is important for design considerations.

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