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Diffusion Welding of MA 6000 and a Conventional Nickel-Base Superalloy

Diffusion welding of superalloys has limited success

BY T. J. MOORE AND T. K. GLASGOW

ABSTRACT. A feasibility study of diffusion welding the oxide dispersion strengthened (ODS) alloy MA 6000 to itself and to conventional Ni-base superalloy Udimet 700 was conducted. Butt joints between MA 6000 pieces and lap joints between Udimet 700 and the ODS alloy were produced by hot pressing for 1.25 hr at temperatures ranging from 1000 to 1200°C (1832–2192°F) in vacuum. Following pressing, all weldments were heat treated and machined into mechanical property test specimens. While three different combinations of recrystallized and unrecrystallized MA 6000 butt joints were produced, the unrecrystallized to unrecrystallized joint was most successful as determined by mechanical properties and microstructural examination. Failure to weld the recrystallized material probably related to a lack of adequate deformation at the weld interface. While recrystallized MA 6000 could be diffusion welded to Udimet 700 in places, complete welding over the entire lap joint was not achieved, again due to the lack of sufficient deformation at the faying surfaces. Several methods are proposed to promote the intimate contact necessary for diffusion welding MA 6000 to itself and to superalloys.

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Introduction

Oxide dispersion strengthened (ODS) Ni-base superalloys, such as MA 6000 (Refs. 1 and 2), are of interest for gas turbine engine applications, because they can possess mechanical properties similar to conventional gamma prime superalloys at low and intermediate temperature regimes (temperatures less than 1000°C/1832°F) as well as the superior strength of simple ODS alloys at more elevated temperature. Unfortunately the intermediate temperature creep rupture strength, creep shear resistance and ductility of ODS Ni-base superalloys are somewhat inferior to those of current superalloys (Refs. 3 and 4); hence the direct substitution of an ODS superalloy for a conventional one is not feasible.

Because gas turbine blades function in a thermal gradient it has been recommended (Ref. 4) that composite turbine blades be considered in which the relatively cool root portion is fabricated from a conventional alloy, and the hotter sections from an ODS alloy such as MA 6000. A typical blading application (Fig. 1) would probably involve two different welds: a lap or shear loaded joint

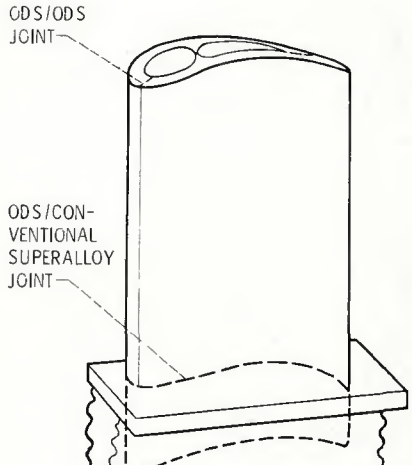

Fig. 1—Conceptual gas turbine blade incorporating oxide dispersion strengthened and conventional nickel-base superalloy materials

between the conventional and ODS alloys near the root section and a longitudinal butt joint between mating ODS alloy pieces to accommodate placement of complex cooling passages. It is envisioned that these materials would have to be joined by diffusion welding since fusion welding would ruin the properties of the ODS material and likely produce cracks in the superalloy. Brazed joints in the ODS alloy, TD-Ni Cr, have exhibited an undesirable tendency to develop pores and thoria-depleted regions as a result of exposure at temperatures ranging from 980 to 1200°C (1796-2192°F) (Ref. 5). Thus, brazed joints in ODS materials are quite unlikely to approach base metal stress rupture strength. To investigate the possibility of fabricating composite blades, a brief investigation of diffusion welding MA 6000 both to itself and to a conventional superalloy was undertaken. The goals were to (1) produce a grain structure at a butt joint in MA 6000 indistinguishable from that of the base metal and a 1000°C (1832°F) joint strength equal to that of the base metal, and (2) match the strength of the weaker material for dissimilar metal joints tested at 760°C (1400°F) in shear parallel to the plane of the weld interface.

Diffusion welding is defined as a solid-state process which involves no macro deformation (Ref. 6). Accordingly, the term diffusion welding is used herein when the welding deformation (decrease in length or thickness) is less than 2 percent. And the term hot pressure welding is used when more than 2 percent deformation was inadvertently produced.

Materials and Procedures

Materials

Commercial MA 6000 has a recrystallized structure which is produced by directional heat treatment. It is an anisotropic material by virtue of the preferred orientation of its large, elongated grains which are a few millimeters in diameter in the short and long transverse bar directions and tens of millimeters long in the longitudinal direction. The orientation has been determined to be [$\bar{1}$11] for the short transverse direction, [$\bar{1}1\bar{2}$] in the long transverse direction and [110] in the longitudinal direction. In this form, MA 6000 shows superior elevated temperature strength in the longitudinal bar direction and somewhat lesser strength in other directions (Ref. 2). As-worked (unrecrystallized) MA 6000, on the other hand, possesses an equiaxed very small grain size, about 0.3 μm in diameter, and no preferred orientation. Elevated temperature testing of unrecrystallized material has shown it to be quite weak, but highly deformable (Ref. 7).

Both recrystallized and unrecrystallized MA 6000 in the form of 27.0 mm wide by 14.3 mm thick by 254 mm long (1.06 × 0.56 × 10 in.) bars were procured from a commercial source. The nominal composition of this material is given in Table 1. Similar information for the superalloy Udimet 700 is also presented in this table. The Udimet 700 material was obtained in the form of wrought 15.1 mm diameter by 1100 mm (0.59 × 43.31 in.) long bars which had been heat treated as follows: 1170°C (2138°F) for 4 hr, A/C; 1080°C (1976°F) for 4 hr, A/C; 845°C (1553°F) for 24 hr, A/C; and finally 760°C (1400°F) for 16 hr, A/C.

Table 1—Chemical Analysis (wt percent) of Base Metals

	Udimet 700	MA6000
C	0.128	–
Cr	15.41	15.5
Co	18.83	–
Mo	5.13	2.0
Al	4.46	4.6
Ti	3.58	2.5
B	0.014	0.02
S	0.005	–
Mn	0.06	–
Si	0.06	–
Fe	0.07	–
Cu	0.04	–
Zr	<0.01	0.10
Ta	–	2.0
W	–	3.5
Y_2O_3	–	1.0
Ni	bal.	bal.

Welding

Method. A vacuum hot press was used to diffusion weld butt joints in MA 6000 and to make dissimilar metal lap joints. Faying surfaces were machined by lapping to a 0.10 μm (4 RMS) finish and degreased in acetone just prior to welding. During heat-up under vacuum, a light pressure (1.2 MPa/174 psi) was maintained on the pieces to be welded. Once at temperature with the vacuum stabilized at 5×10^{-5} torr or lower, the full welding force was applied and held for 1.25 hr. Following completion of this cycle, the power to the furnace was discontinued and the welding pressure was maintained as the weldment was allowed to cool under vacuum.

Butt Joints. In keeping with the blade geometry shown in Fig. 1 and the directional mechanical properties of the ODS alloy, diffusion welds of MA 6000 to itself were made with the faying surface parallel to the plane described by the short transverse and longitudinal bar directions. A schematic drawing of an as-welded block assembly and its relationship to the original MA 6000 bars is shown in Fig. 2A. Approximately six mechanical property test specimens (Fig. 2B) were machined from each block with the weld interface located within the gage section. Diffusion weldments were made with MA 6000 in three different starting conditions: Recrystallized/Recrystallized, Recrystallized/Unrecrystallized and Unrecrystallized/Unrecrystallized. A summary of the solid-state welding parameters and aver-

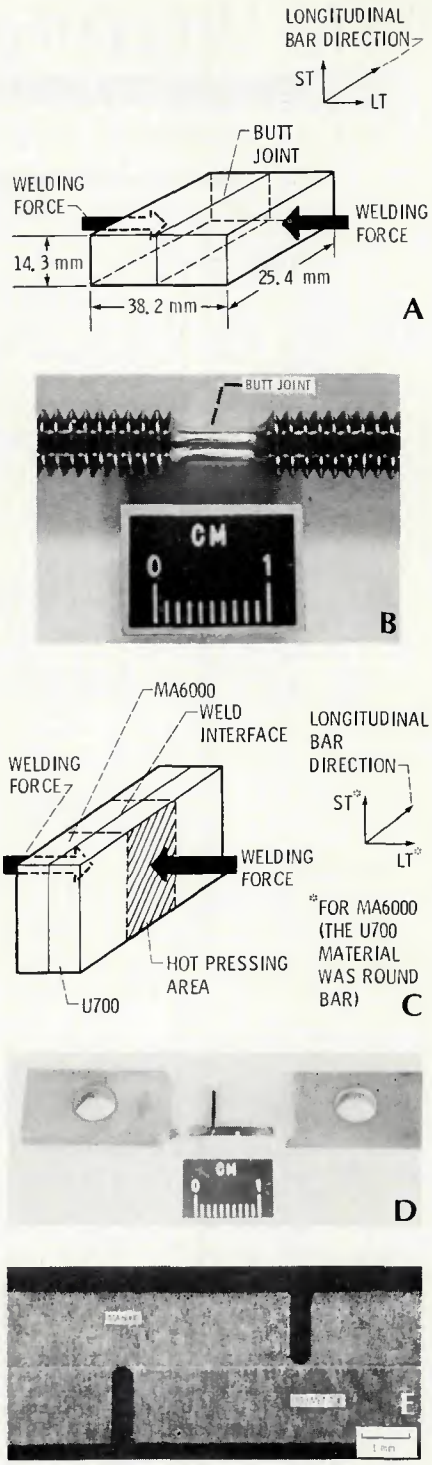

Fig. 2—(A) Butt joint configuration—(B) Tensile test specimen—(C) Lap joint geometry—(D) Shear test specimen—(E) Notched region of shear specimen

Fig. 3—Gross cracking in recrystallized MA 6000 hot pressure welded at 138 MPa at 1200°C for 2 hr. Total reduction in thickness was 22 percent

age deformation noted in the welding direction (decrease in length) are given in Table 2. All butt joint block weldments were post weld heat treated in air with the following schedule: 1 hr at 1250°C (2282°F), A/C; 2 hr at 955°C (1751°F), A/C; and finally 24 hr at 845°C (1553°F), A/C.

Lap Joints. Also in keeping with the geometry of Fig. 1, lap joint welds were made between the ODS and the conventional superalloy with the faying surfaces being the short transverse-longitudinal plane for MA 6000 and bar diameter-longitudinal plane for U-700. Welds were made by hot pressing the center sections of 1.6 mm thick, 12.7 mm wide, 64 mm long (0.06 × 0.50 × 2.52 in.) pieces of recrystallized MA 6000 to a similar piece of superalloy, as shown in Fig. 2C. A listing of the diffusion welding variables and deformation observed in the hot pressing (thickness) direction is given in Table 2. Following post weld heat treatment of 4 hr at 1175°C (2147°F), A/C; 4 hr at 1080°C (1976°F), A/C; 24 hr at 840°C (1544°F), A/C and finally 16 hr at 760°C (1400°F), A/C, a notched tensile shear specimen (Fig. 2D) was cut from each weldment where the depth and location of the notches insured shear loading at the weld interface with an overlap of two times the thickness—Fig. 2E.

Results

Welding

While diffusion welding of recrystallized MA 6000 to itself would be ideal

Table 2—Welding Variables (Welding time 1.25 h in a 5×10^{-5} torr or better vacuum.)

Materials combination	Number of weldments	Temp., °C	Pressure, MPa	Average deformation, percent
		MA 6000 Butt Joints		
Unrx/Unrx	2	1000	27.6	1.6
Rx/Unrx	1	1200	2.1	0.7
Rx/Unrx	1	1200	8.3	9.1
Rx/Unrx	3	1000	35.0	1.0
Rx/Unrx	1	1000	41.4	1.8
Rx/Rx	1	1200	69.0	0.1
Rx/Rx	2	1200	86.2	0.1
Rx/Rx	2	1200	103.0	0.1
		MA 6000/Udimet 700 Lap Joints		
Rx/U700	5	1120	34.5	0.2
Rx/U700	7	1100	34.5	0.1

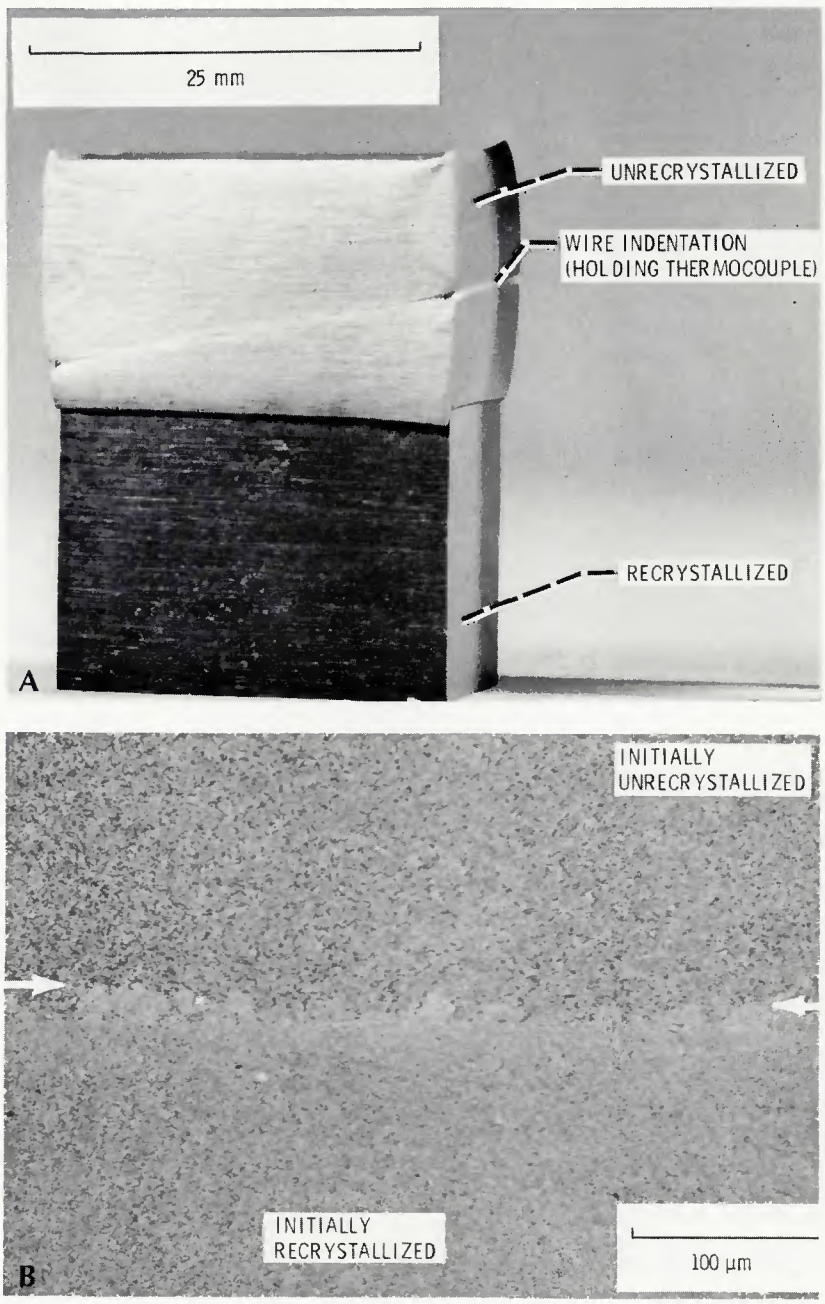

Fig. 4—Hot pressure weld of unrecrystallized MA 6000 to recrystallized MA 6000 at 1200°C and 8.3 MPa for 1.25 hr. Overall change in length was 9 percent; however, all deformation occurred in the unrecrystallized material

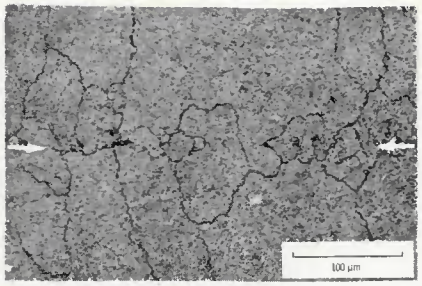

Fig. 5—Post weld heat treated microstructure at the weld joint of initially unrecrystallized MA 6000 diffusion welded to itself at 1000°C and 27.6 MPa for 1.25 hr. Overall deformation was 1.6 percent

since such welds could retain as received mechanical properties, this type of welding appears to be impractical. Preliminary experiments revealed that very high stresses were required to produce the deformation necessary for welding. Unfortunately, the transition between true plastic flow and massive cracking exists over a very narrow range of stress. Figure 3 illustrates the unacceptable crushing encountered in the hot pressure welding of recrystallized MA 6000 at 1200°C (2192°F) and 138 MPa (20,010 psi). Subsequently, five MA 6000 diffusion welded butt joints were made at 1200°C (2192°F) at pressures of 69.0 (10,005 psi), 86.2 (12,499 psi), and 103.0 MPa (14,935 psi)—Table 2. For these weldments, which were made with 0.1 percent deformation, no cracking was observed. However, all five welds were of poor quality and failures occurred at the joints during handling or during machining.

Since the recrystallized MA 6000 is too creep resistant to be welded to itself, it was felt that the joining of unrecrystallized to recrystallized material would be possible due to plastic flow in the unrecrystallized MA 6000, as shown in Fig. 4A. While welding at 1200°C (2192°F) could produce a sound weld (Fig. 4B and Table 2), such processing resulted in partial recrystallization of the initially unrecrystallized material as revealed by hardness measurements prior to and after welding. In general, recrystallization during welding of MA 6000 is to be avoided, since it leads to a small grain size such as appears near the weld interface in Fig. 4B and inferior high temperature mechanical properties (Ref. 8). Therefore, further hot press runs were conducted at 1000°C (1832°F) (Table 2) where hardness testing before and after welding gave little evidence of recrystallization. Unfortunately, none of the 1000°C (1832°F) diffusion welds were sound, even after undergoing up to 2 percent deformation, as all test bars failed at the joint during machining. Apparently, more than 2 percent deformation is required at 1000°C (1832°F) in order to produce the intimate contact required for complete solid-state welding since all of the deformation takes place in the unrecrystallized material.

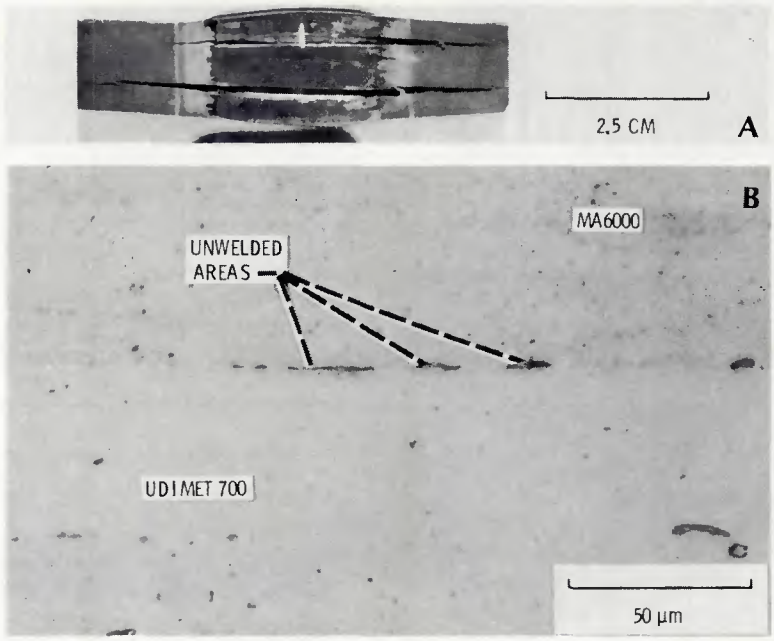

Fig. 6—Dissimilar metal lap weldments between recrystallized MA 6000 and Udimet 700—(A) Large scale cracking in the MA 6000 after hot pressure welding at 1180°C and 34.5 MPa for 1.25 hr with 12.5 percent deformation—(B) Typical unwelded areas in a post weld heat treated joint diffusion welded at 1100°C and 34.5 MPa for 1.25 hr with 0.1 percent deformation

In light of the above difficulty in the welding of recrystallized to unrecrystallized MA 6000 at 1000°C (1832°F), the apparent ease of diffusion welding the unrecrystallized material to itself under similar conditions was surprising. The welds produced by 1000°C (1832°F), low stress hot pressing (Table 2) were microstructurally sound (Fig. 5) as evidenced by no unwelded regions and grain growth across the original interface.

In preliminary attempts to hot press weld recrystallized MA 6000 to Udimet 700 at 1180°C (2156°F), MA 6000 was subject to the same cracking problems (Fig. 6A) which were encountered in the self welding of the ODS recrystallized material—Fig. 3. For a constant welding pressure of 34.5 MPa (5002 psi), the extent of cracking was reduced as the welding temperature was lowered. At either 1120 or 1100°C (2048 or 2012°F), crack free diffusion welds were produced (Table 2), although the measured deformation was quite small and unwelded regions were found—Fig. 6B. In addition, all lap joint specimens had a slight bow, with the U-700 on the concave side, apparently due to slight differences in thermal expansion between the two materials.

Mechanical Properties

MA 6000/MA 6000. Mechanical property tests could not be conducted on recrystallized material welded to itself or on recrystallized to unrecrystallized

Table 3—1000°C Tensile Test Data for MA 6000 Butt Joints Tested in the Post Weld Heat Treated Condition at a Constant Crosshead Speed of 0.0042 mm/s

Welding Parameters				Test Results		
Materials combination	Temp., °C	Stress, MPa	Percent deformation	Ult. Tens. Str., MPa	%RA	Fracture
Unrx/Unrx	1000	27.6	1.3	294	0.0	Base metal, intergranular
Rx/Unrx	1200	2.1	0.7	53.0	63.9	Unrx base metal, cup and cone
Rx/Unrx	1200	2.1	0.7	53.1	0.0	At/near joint, 30% smooth, 70% intergranular
Rx/Unrx	1200	8.3	9.1	67.4	0.0	Near joint, intergranular
Rx/Unrx	1200	8.3	9.1	67.3	0.0	At/near joint, 30% smooth, 70% intergranular

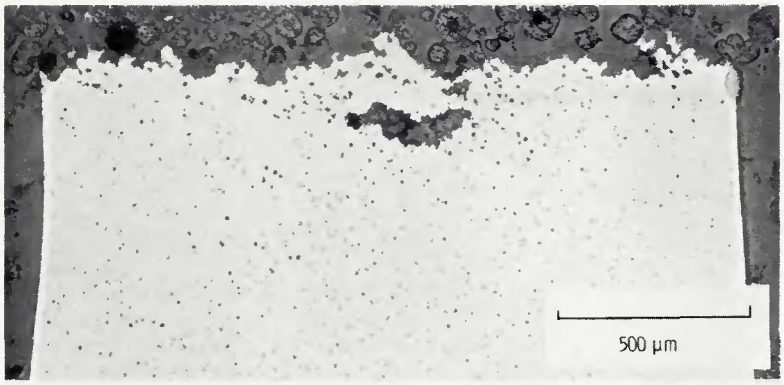

Fig. 7—Microstructure in the vicinity of the 1000°C tensile fracture region of post weld heat treated unrecrystallized MA 6000 diffusion welded to itself at 1000°C and 27.6 MPa for 1.25 hr with 1.3 percent deformation. The ultimate tensile strength was 294 MPa

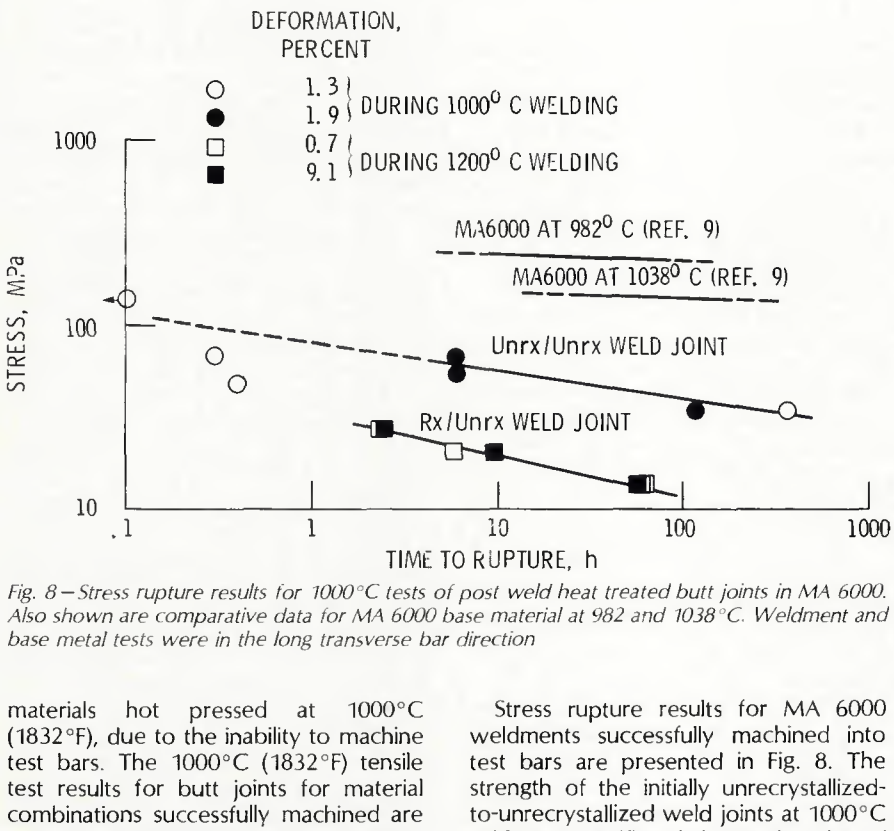

Fig. 8—Stress rupture results for 1000°C tests of post weld heat treated butt joints in MA 6000. Also shown are comparative data for MA 6000 base material at 982 and 1038°C. Weldment and base metal tests were in the long transverse bar direction

materials hot pressed at 1000°C (1832°F), due to the inability to machine test bars. The 1000°C (1832°F) tensile test results for butt joints for material combinations successfully machined are presented in Table 3. The tensile strength of the initially unrecrystallized to recrystallized material combination was on the order of 60 MPa (8700 psi) for both welding conditions, and failure was generally intergranular near the weld region with little evidence of plastic flow. However, in one case extensive deformation occurred with final failure in the initially unrecrystallized material. The single test result for the self welded unrecrystallized material gave a tensile strength of 294 MPa (42,630 psi) with intergranular failure in the base metal as shown in Fig. 7. This value is well below the transverse tensile strength (about 338 MPa/49,010 psi) of commercial MA 6000 which is recrystallized by directional heat treatment (Ref. 9).

Stress rupture results for MA 6000 weldments successfully machined into test bars are presented in Fig. 8. The strength of the initially unrecrystallized-to-unrecrystallized weld joints at 1000°C (1832°F) is significantly better than that of the recrystallized to unrecrystallized joints. For the latter combination, essentially all specimens failed near the weld joint in a weaker, small grained area of the initially unrecrystallized material by intergranular fracture with little if any measurable reduction in area. No dependence on the amount of prior deformation during welding was evident. On the other hand, test bars machined from the initially unrecrystallized to unrecrystallized weld joints did indicate a dependency on the amount of plastic flow during hot pressing where the higher strain yielded more consistently, stronger welds. Although most of these stress rupture specimens failed at or near the weld region and little deformation on

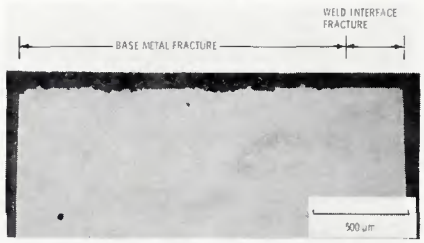

Fig. 9—Partial failure at weld interface of a stress rupture test bar machined from unrecrystallized MA 6000 diffusion welded to itself at 1000°C and 27.6 MPa for 1.25 hr with 1.3 percent deformation during welding. This specimen was stress rupture tested at 1000°C and 69 MPa for 0.3 hr in the post weld heat treated condition

failure was evident, the bars subjected to about 1.9 percent strain during welding exhibited intergranular base metal fracture while the bars produced with around 1.3 percent strain during hot pressing generally displayed significant regions of failure along the weld interface—Fig. 9.

MA 6000/Udimet 700. The results of the shear tests conducted at 760°C (1400°F) in a universal testing machine are given in Table 4. In each case, tensile type fracture occurred in the MA 6000 generally away from the lap joint, as shown in Fig. 10A. Since the calculated tensile stresses at fracture, 395 to 568 MPa (57,275–82,360 psi), were considerably less than the 944 MPa (136,880 psi) ultimate tensile strength of MA 6000 (Ref. 9), bending during testing apparently reduced the strength of the MA 6000 base metal.

While one shear rupture test specimen exhibited the fast strain rate type fracture mode, all other long term, constant load tests failed at the lap joint. The fracture surfaces in Figs. 10B and C are typical of the latter behavior where regions of sound welds, as evidenced by the torn, pulled out MA 6000, and regions of no welding, as witnessed by machining scratches are both visible. While the dissimilar metal lap joints are, admittedly, not of good quality, the shear rupture data obtained in this study is plotted in Fig. 11 to illustrate that some strength can

Table 4—Shear Strength Results for Post Weld Heat Treated MA 6000/Udimet 700 Lap Joints Tested at 760° C and a Constant Crosshead Speed of 0.0042 mm/s

Welding Variables		Calculated Stress at fracture, MPa	
Temp., °C	Deformation, percent	Shear	Tensile
1120[a]	0.07	283	568
1100[a]	0.15	230	461
1100[a]	0.08	199	395

(a) 34.5 MPa for 1.25 h

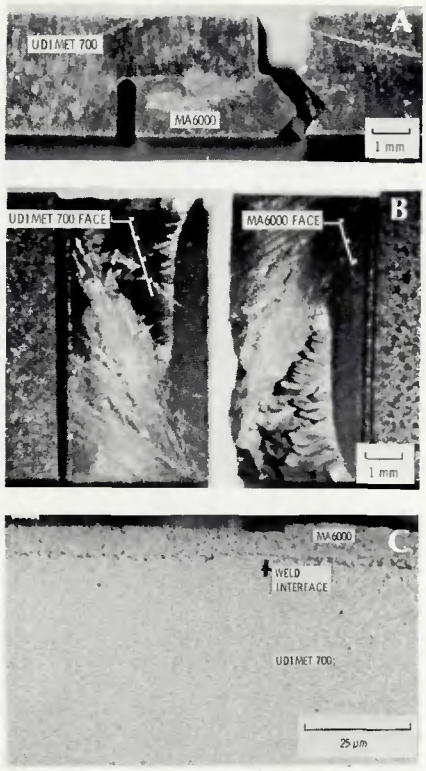

Fig. 10—Failure in MA 6000/Udimet 700 lap joint shear specimens tested at 760°C—(A) Tensile fracture in MA 6000 after shear testing at a crosshead speed of 0.0042 mm/s—(B) Shear faces of a shear rupture specimen tested at 124 MPa for 9.1 hr—(C) Microstructure of shear failure after shear rupture testing at 117 MPa for 100 hr

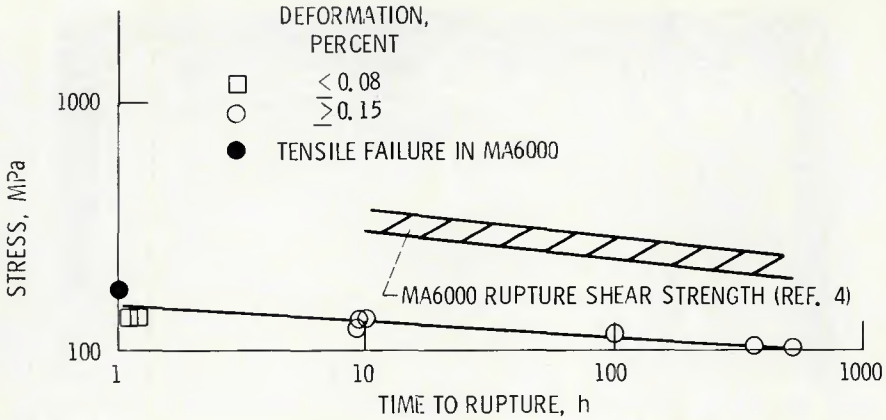

Fig. 11—Shear rupture test results for MA 6000/Udimet 700 lap joints tested at 760°C after a post weld heat treatment. Also shown are comparative data for MA 6000 base metal

be obtained even with incomplete welds.

Discussion

This study demonstrated that MA 6000 can be diffusion welded to itself by hot pressing when unrecrystallized MA 6000 is utilized. While the welds were sound and grain growth occurred across the weld interface, some small grains were present—Fig. 5. Such a structure is almost always detrimental to high temperature strength in ODS alloys (Refs. 8 and 10). The heat treatment employed in this study was not the same as the thermal gradient heat treatment employed to produce the oriented, large grain structure, strong recrystallized material. This lack of sophistication yielded a grain size and shape which is quite different from the recrystallized structure produced by directional heat treatment. These two factors lead to the mechanical properties for butt joints (Table 3 and Fig. 8) which are less than those of the recrystallized MA 6000.

Neither of the above two problems concerning the diffusion welding of unrecrystallized MA 6000 should be intractable. Previous welding studies (Refs. 8, 10, and 11) of ODS alloys demonstrated that recrystallization of small grains at the joint could be minimized by chemical polishing or by electropolishing prior to welding. Additionally, the as-welded unrecrystallized blade shape could be given the same commercial thermal gradient heat treatment as normal extruded plus hot rolled MA 6000. Special care must, however, be exercised during diffusion welding in order that sufficient deformation, about 2 percent, occurs to insure intimate contact while maintaining the welding temperature and strain below levels required to trigger recrystallization during welding or premature recrystallization during heat treatment.

The diffusion welding of recrystallized MA 6000 to itself or unrecrystallized to recrystallized materials does not appear to be practical, due to the inability of the recrystallized stock to plastically flow. This, however, could be due to the temperatures and deformation rates used in this work. As pointed out in the Appendix, it may be possible to weld recrystallized MA 6000 at lower welding temperatures with a fast enforced deformation rate. Complete grain growth across the weld interface could be difficult to achieve, however, when the starting material is in the recrystallized condition (Ref. 8).

Shear rupture testing (Fig. 11) and post test structure (Figs. 10B and C) of lap joints between U-700 and recrystallized MA 6000 revealed that it is possible to produce diffusion welds between super and ODS alloys. Unfortunately, the inability to produce complete welding over the entire lap joint resulted in shear properties significantly less than those for MA 6000—Fig. 11. The lack of welding is probably related to the creep resistance of the recrystallized MA 6000 component, which prevented sufficient plastic flow at the interface necessary to promote intimate contact everywhere. Two solutions to this problem seem possible:

1. A soft, thin Ni-base interlayer could be applied to one faying surface prior to diffusion welding. Use of such an interlayer should produce a sound weld joint, and because of the high temperatures involved in pressing and heat treatment, sufficient diffusion would occur to change the composition of the interlayer to that of a superalloy. Owczarski, et al. (Ref. 12) used a Ni-30 Co interlayer to produce joints in Udimet 700 that were as strong as the base metal in 982°C (1800°F) stress rupture tests.

2. As was noted above, the compressive ductility of commercial MA 6000 increases when the temperature is decreased and strain rate is increased; thus, it could be possible to produce sufficient deformation for complete welding at the faying surfaces by hot pressing at lower temperatures (<1100°C/<2012°F) and forcing flow by higher pressures (>34.5 MPa/>5003 psi).

The results from the dissimilar metal lap joint testing indicate several areas of concern. Tensile fracture at relatively low strength levels in the MA 6000 portion of the test specimen during fast strain rate shear testing (Table 4 and Fig. 10) could be indicative of a poor specimen design which leads to bending or some other factor and premature failure. Comparison of the fast strain rate and the shear rupture results demonstrate that fast testing does not adequately evaluate the quality of the shear joint. Finally, examination of the lap joints showed that in the regions of good welding, shear rupture failure only occurred in the MA 6000. This result reveals that the shear strength of MA 6000 will be controlling and it is not weak because of small grains since these MA 6000 coupons had the recrystallized structure produced by directional heat treatment. Hence, the transition joint between superalloy and ODS alloy must be located in a region where the shear strength of MA 6000 is sufficient to support all envisioned states of stress.

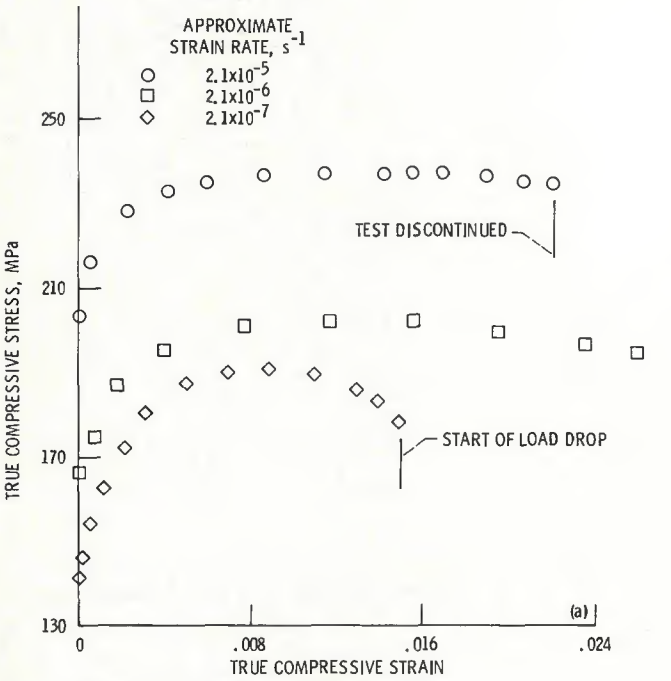

Fig. A-1a — True compressive strain diagram

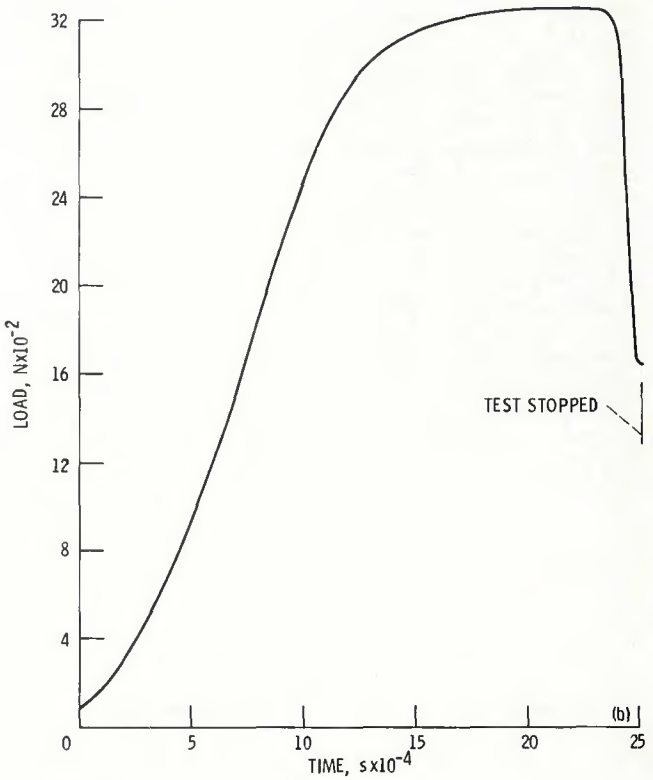

Fig. A-1b — Load/time curve

Summary of Results

A feasibility study of hot pressing to produce butt joint diffusion welds of ODS MA 6000 to itself and lap joint diffusion welds between MA 6000 and the conventional superalloy Udimet 700 has been undertaken. This study has demonstrated:

1. MA 6000 can be successfully welded to itself at 1000°C (1832°F) when unrecrystallized materials are used as evidenced by base metal failure in 1000°C (1832°F) stress rupture tests.

2. Attempts to make strong butt joints in recrystallized MA 6000 were unsuccessful. Butt joints made between recrystallized and unrecrystallized materials at 1200°C (2192°F) were weaker in 1000°C (1832°F) stress rupture tests than the unrecrystallized/unrecrystallized joints.

3. Diffusion welding of MA 6000 to Udimet 700 is possible; however, under the conditions investigated, complete welding of the faying surfaces was not accomplished and the joint was relatively weak in 760°C (1400°F) shear rupture tests.

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Appendix

Elevated Temperature Compressive Ductility in the Oxide Dispersion Strengthened Alloy MA 6000

BY J. DANIEL WHITTENBERGER

As part of a study (Ref. A1) of deformation and fracture in the recrystallized oxide dispersion strengthened (ODS) alloy MA 6000, constant velocity compression tests were performed on 5 mm (0.2 in.) diameter by 10 mm (0.4 in.) long cylindrical test specimens between 871 and 1093°C (1600–1999°F) in air. Specimens were taken with their lengths parallel to the longitudinal, long transverse and short transverse bar directions which corresponded to [110], [$\bar{1}1\bar{2}$] and [$\bar{1}11$] directions since the alloy bar had a strong preferred orientation. The nominal composition of the MA 6000 was 15Cr-2Mo-4W-2Ta-4.5Al-2.5Ti-.15Zr-.01B-.05C-1.1Y_2O_3-balance Ni. Grain width and thickness in the long transverse and short transverse bar directions were on the order of several millimeters, and grain lengths in the longitudinal bar direction were on the order of several centimeters.

Typical true compressive stress-true compressive strain diagrams are presented in Fig. A-1a where the stress rapidly rises to an approximately constant value, and after some flow at this level additional strain occurs at decreasing stress values. While the stress required for continued deformation slowly decreased once through the constant flow stress regime for the majority of the tests, many longitudinal and short transverse specimens tested at the more elevated temperatures and slower rates possessed large (30 to 50 percent) drops in the load-time curves, as illustrated in Fig. A-1b. The slow decrease in flow stress is believed to be due to concentrated slip in shear bands, and the large load drops are apparently the result of crack formation and propagation in such bands (Ref. A1).

Examination of all the true stress-true strain data revealed that the strain range for homogeneous deformation (uniform plastic strain throughout the test specimen) was, in general, independent of test direction but decreased as the test temperature was increased or strain rate was decreased. This behavior is illustrated in Fig. A2 for the average strain range (averaged over the three orientations) for

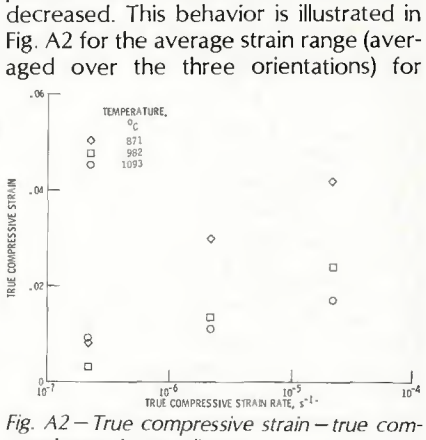

Fig. A2 — True compressive strain — true compressive strain rate diagram

which the flow stress was approximately constant. Clearly, the available compressive ductility in MA 6000 changes in a manner opposite to normal expectations.

The general behavior that the amount of uniform deformation increases as the strain rate is increased and/or temperature is decreased could have important implications with regard to the welding of MA 6000. Moore and Glasgow (main paper) found that recrystallized MA 6000 could not be successfully solid-state welded at 1200°C (2192°F) due to the inability to produce reasonable, uniform deformation at the interface. Figure A-2 indicates that such deformation might be possible under fast strain rate and/or reduced temperature conditions.

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