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Methods for Diffusion Welding the Superalloy Udimet 700

High quality welds can be produced in cast and wrought materials by using a thin electroplated Ni-Co interlayer at the mating surfaces and carefully controlling bonding parameters

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ABSTRACT. The diffusion-welding characteristics of wrought and cast Udimet 700 were studied to develop methods of producing high quality, strong diffusion-welded joints in high-strength nickel-base superalloys. It was found initially that a number of factors inhibited the formation of satisfactory diffusion welds. The most serious of these was the formation during bonding of a continuous array of stable Ti(C,N) and $NiTiO_3$ precipitates at the joint interface. These produced a weak, planar joint grain boundary causing very poor joint properties. Electroplating a thin nickel interlayer on each of the mating surfaces prior to welding successfully prevented the formation of the precipitates and accelerated elimination of joint porosity because of the interlayer's softness. However, nickel interlayers produced some residual chemical and microstructural heterogeneity at the joint.

Even greater improvements in joint quality were achieved by utilizing a Ni-Co alloy electroplate as the interlayer material. This resulted in better chemical and microstructural homogenization at the joint and reduced the sensitivity of joint quality to slight variations in process parameters. Concurrently, improved welding parameters were developed which more effectively migrated the joint grain boundary to the desired non-planar configuration. When welding was conducted with the improved process parameters and Ni-Co interlayers, strong void-free joints were produced in both wrought and cast Udimet 700. Limited testing indicated that the properties of diffusion welds in the wrought alloy approached those of the base metal in room temperature and 1400F tensile tests. Cast alloy diffusion welds were made which exhibited $\geq$90% joint efficiencies in creep-rupture tests at 1400F and 1800F.

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Introduction

The performance characteristics demanded of advanced aircraft gas turbine engines have led to the extensive use of air-cooled components. However, the complex configurations of some of these turbine parts have proven difficult and expensive to produce by precision casting techniques. One solution to these problems would be to produce the components in economical, readily fabricable segments and subsequently join them to form the desired configuration. A joining process which is promising for many of these components is diffusion welding.

Diffusion welding has two significant advantages: 1) it allows joining of complex configurations without introducing melting or cracking and without unpredictable distortion; and 2) the properties and microstructure of the resultant joint can be equivalant to that of the base material being joined. Previous studies[1, 2] have defined the basic mechanisms associated with diffusion welding and have demonstrated the feasibility of this process as a method of joining titanium alloys. However, many of the potential applications for diffusion welding in the manufacture of gas turbines would involve the joining of high-strength nickel-base superalloys.

Only limited work has been done on the diffusion welding of precipitation-hardenable nickel-base superalloys.[3, 4] While these prior studies revealed some of the problems of diffusion welding nickel-base superalloys, the welding techniques produced joints with elevated temperature mechanical properties which were lower than those of the base metals. Furthermore, the alloys which were examined had considerably lower Al + Ti contents than do the alloys used in advanced turbine components. Aluminum and titanium contents significantly affect the diffusion welding characteristics of nickel-base alloys.[4]

Therefore, this detailed study was undertaken to examine the diffusion welding characteristics of a representative high-strength nickel-base superalloy and to ascertain if high-strength bonds could be produced in this type

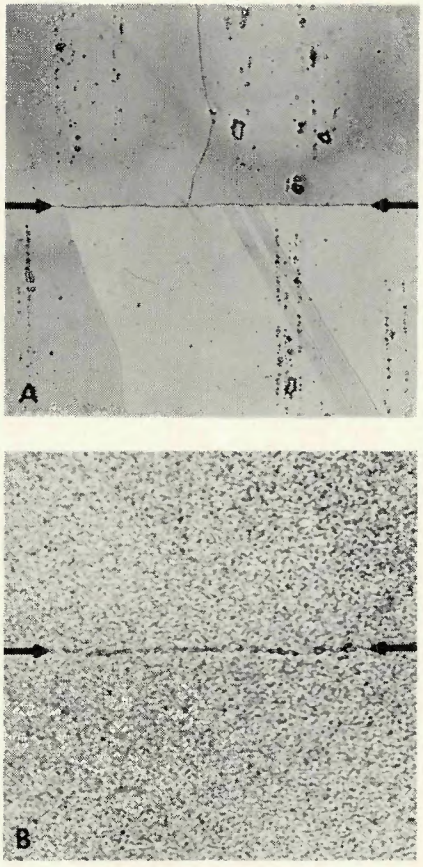

Fig. 1—Microstructure of wrought Udimet 700 sample welded without an interlayer at 2140F/4 hours/1 ksi. Note the precipitates at the immobilized, planar interface boundary. Mag: (A) 100X, (B) 1000X

Table 1—Chemical Compositions of Udimet 700

Element	Composition, wt. % Wrought	Cast
Nickel	Balance	Balance
Chromium	15.1	14.7
Molybdenum	4.95	4.15
Cobalt	18.7	15.2
Aluminum	4.49	4.32
Titanium	3.44	3.43
Carbon	0.06	0.07
Boron	0.014	0.014
Zirconium	< 0.05	< 0.04
Iron	0.15	0.10
Manganese	< 0.10	< 0.1
Silicon	< 0.10	< 0.1
Sulfur	0.005	0.003

of material. Udimet 700 was selected for the study since methods were sought to satisfactorily diffusion weld both wrought and cast forms of a high-strength superalloy.

Materials and Procedures

Specimens of wrought Udimet 700 were machined from $^5/_8$ in. diam hot-rolled barstock while the cast samples were taken from $^5/_8$ in. diam by 6 in. long as-cast bars. The chemical compositions of these heats of material are listed in Table 1. The wrought specimens were bonded in either the as-received (mill-annealed) condition or after solution heat treatment while the cast specimens were joined in either the as-cast or a solution heat treated condition. It was found that the initial condition of the wrought and cast samples had no observable effect on the diffusion welding characteristics. Most samples were heat treated after diffusion welding. Various heat treatments were employed after welding and these are described during the discussion of specific specimens in the *Results* section. However, a common procedure was to utilize the welding cycle as the sample's solution heat treatment. After welding, the samples were given the post-weld aging heat treatment described in Table 2. Typical solution and post-weld aging heat treatments are also listed in the table.

All diffusion welded samples were made by butt welding cylindrical specimens which were induction heated in a vacuum ($<10^{-5}$ torr) while under a compressive load. The apparatus used for this work has been described in detail previously.[1] Diffusion welds prepared for metallographic examination were made from two 0.6 in. diam by 0.75 in. long cylinders while each mechanical-properties specimen was machined from two 0.6 in. diam by 3 in. long cylinders butt welded together. The ends of the cylinders were ground flat and parallel, and then the ends to be joined were polished with a 600-grit metallographic paper.

The mating surfaces of some specimens were subsequently electroplated as will be described in the *Results* section. Immediately prior to welding, the mating surfaces were degreased successively with acetone and methanol, rinsed with demineralized water, and then blown dry with Freon.

After welding and appropriate heat treatments, the metallographic specimens were prepared by sectioning in an axial direction normal to the diffusion weld. The samples were mechanically polished in a conventional manner and then immersion etched in a 30 parts lactic acid, 20 parts hydrochloric acid, 10 parts nitric acid, 5 parts 30% hydrogen peroxide solution. The samples were examined by light and electron metallography and in some cases with an electron microprobe.

Results

Metallographic Study

A preliminary parametric survey was first made to determine the range of bonding temperatures, pressures and times which produced satisfactory diffusion welds in wrought and cast Udimet 700. The initial diffusion welds were made on both wrought and cast specimens without any interlayer placed at the mating surfaces. An example of one such bond in the wrought alloy is shown in Fig. 1. Essentially 100% metallurgical bonding has occurred, i.e., no unbonded areas or even microporosity were detectable at the joint.* However, it can be seen that the interfacial boundary remained planar (even though other grain boundaries in the bulk of the wrought specimen had migrated extensively). Close metallographic examination of the interfacial grain boundaries in both the wrought and cast specimens revealed that they were pinned by a row of fine precipitates (e.g., Fig. 1b).

It has frequently been hypothesized that oxides of aluminum would hinder the bonding of high strength superalloys because of the relatively large amounts of this element in these materials. However, electron probe microanalysis of these interfacial precipitates showed that they were rich in titanium rather than aluminum. The precipitates at the joint in the sample in Fig. 1 were extracted from the metallographic specimen onto a replica after etching with a 2% HCl-in-methanol solution. Extraction replicas were also taken from the fracture surface of a similar specimen which

Table 2—Typical Heat Treatments Employed in This Study

Wrought Udimet 700	Heat Treatment*	Cast Udimet 700
2140F/4 hr/forced air cool (FAC)	Solution	2125F/2 hr/forced air cool (FAC)
1975F/4 hr/FAC +	Postweld	2125F/15 min → cool at
1550F/4 hr/FAC +	Aging	≃ 100F/hr to 1975F/4 hr/
1400F/16 hr/FAC	treatment	FAC + 1400F/16 hr/FAC

* All heat treatments conducted in argon

* As is discussed later, the welding parameters utilized in this study consistently produced complete bonding in relatively short times for a variety of sample and mating-surface conditions. The elimination of joint porosity turned out to be only a minor problem while the major difficulty associated with the diffusion welding of superalloys was to achieve satisfactory joint homogenization and interfacial boundary migration.

was broken through the joint after bonding for 1 hour. From electron-diffraction analysis of these replicas, the interfacial precipitates were identified as Ti(C, N) and $NiTiO_3$ particles. (The "d" spacings obtained from electron diffraction are listed in Table 3.)

The interfacial precipitates which formed during diffusion welding of the wrought alloy were uniformly distributed across the joint (Fig. 1). In diffusion bonds of the cast alloy, however, the interfacial precipitate distribution was more heterogeneous. There were heavy concentrations of the interface precipitate at locations which were bordered by titanium-rich, interdendritic regions of the base metal. Where the joint interface was bounded by titanium-lean dendritic core areas, there was less interface precipitate. These observations indicated that the formation of the interfacial precipitates was dependent upon the local concentrations of base metal elements, particularly titanium, at the mating surfaces.

The formation of the Ti(C,N) and $NiTiO_3$ interfacial particles is attributed to a reaction between elements in the alloy (e.g., titanium) and residual surface contaminants. Although the mating surfaces were carefully degreased and dried prior to bonding in a vacuum, some form of surface contamination is retained as an adsorbed layer. Detailed analysis of surface compositions by Auger spectroscopy has shown that carbon and oxygen are frequent surface contaminants even on freshly broken samples exposed to ultra-high vacuum, e.g. 10^{-9} torr.[5]

Experiments with Udimet 700 samples whose mating surfaces had been electropolished indicated that the amount of interface precipitation was reduced somewhat both by this surface treatment and by rapidly placing the samples under vacuum rather than exposing the freshly prepared surfaces to the room environment for 5 to 15 minutes prior to bonding. However, it was apparent that use of such techniques to eliminate interfacial precipitation would be both unreliable and impractical in the production diffusion welding of actual hardware.

It was found, as expected, that the diffusion welds containing a precipitate-laden, planar joint interface had poor mechanical properties. Consequently, an effort was made to find a practical means of preventing their formation and promoting the movement of the interfacial grain boundary. A nickel interlayer produced by electroplating each mating surface (using a nickel sulfamate bath) was found to be beneficial. The electroplated nickel at the interface separated the surface contaminants from the titanium in the base metal and diluted their concentration such that the deleterious precipitates were prevented from forming during bonding.

Table 3—Electron Diffraction Analysis of Interfacial Precipitates Formed During Diffusion Welding of Wrought Udimet 700 Without Interlayer Material

"A"-Type Particle

Electron diffraction	TiC (ASTM 6-0614)		TiN (ASTM 6-0642)		
d (Å)	d (Å)	I/I_o	d (Å)	I/I_o	hkℓ
2.48	2.51	80	2.44	75	111
2.16	2.179	100	2.12	100	200
1.48	1.535	50	1.496	55	220
1.27	1.311	30	1.277	25	311
1.22	1.255	10	1.223	16	222
—	1.086	5	1.059	8	400
—	0.997	5	0.972	12	331
0.943	0.971	30	0.948	20	420

"B-Type" Particle

Electron diffraction	$NiTiO_3$ (ASTM 17-617)		
d (Å)	d (Å)	I/I_o	hkℓ
—	4.61	5	003
3.62	3.69	40	012
2.69	2.710	100	104
2.53	2.518	80	110
—	2.331	5	015
—	2.300	5	006
2.20	2.210	30	113
—	2.154	5	021
2.08	2.078	10	202
1.84	1.843	50	024
1.70	1.698	80	116
1.64	1.602	20	018, 122
1.48	1.487	40	214
1.46	1.452	40	030

Samples were also welded using a similar-thickness nickel foil rather than an electroplate as the interlayer but it was found that some precipitates still formed at the foil-to-base metal interface. (This sandwich configuration allowed direct contact between the base metal surface and adsorbed contaminants because of the exposure of this surface to the room environment prior to assembly of the sample for bonding.) Thus subsequent welds were made using electroplated interlayers.

By studying the effect of interlayer thickness, it was found that a total of 5 microns of nickel (2.5 microns on each mating half) was optimum. At this thickness there was essentially no precipitate formation during diffusion welding and the interfacial grain boundaries in the wrought specimens were able to migrate freely out of the original joint plane. Thicker interlayers increased the tendency for chemical and microstructural heterogeneity which remained at the joint after the welding cycle and full post-weld heat treatment. Joints made with thinner nickel interlayers (e.g., 2 microns total) failed to effectively suppress the interfacial precipitate reaction.

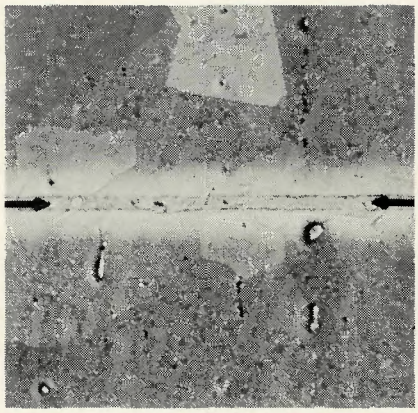

Fig. 2—Diffusion weld made in nickel-plated wrought Udimet 700 by bonding for 5 minutes at 2140F and 1 ksi. Lack of visible porosity was typical of short-time welds in samples bonded with a nickel interlayer. Mag: 250X

The use of a nickel interlayer at the mating surfaces also accelerated elimination of porosity during the early stages of the bonding process. Figure 2 shows a diffusion weld using the nickel interlayer which was made in 5 minutes at 2140F and 1 ksi in the wrought alloy. No porosity is visible after this short bonding time. This specimen and one welded for 5 minutes without an interlayer were subsequently fractured through their joints by impacting at room temperature.

The fractured surfaces of these samples are shown in Fig. 3. Although both specimens failed through the joint, the sample with the nickel interlayer was much more difficult to break and failed with much greater local deformation (Fig. 3b) than did the specimen without an interlayer (Fig. 3a). On the sample made without an interlayer, grooves (from the 600-grit surface preparation) which had not yet been deformed to contact with the mating surface are visible (Fig. 3a). Similar grooves were not observed on the nickel plated sample because the soft nickel interlayer quickly conformed at the bonding temperature.

While the use of a thin nickel interlayer resulted in precipitate-free, low porosity joints after short welding times (Fig. 2), it was found that joint homogenization and migration of the interfacial boundaries out of the original joint plane were also necessary if good mechanical properties were to be achieved. A high degree of chemical homogenization was desired across the location of the original nickel interlayer so that this region would be uniformly strengthened by γ' (Ni_3Al, Ti) precipitation (to a degree comparable to the base metal) during post-weld aging. To achieve the required diffusional homogenization, it was necessary to expose the joint area to high temperature for times considerably longer (4 hours) than that required to just produce a void-free diffusion weld (10 minutes).

It was also found necessary to weld at a temperature above the solution temperature of the strengthening γ' precipitate. When temperatures below the γ-γ' solvus were employed, a near continuous layer of γ' formed at the joint interface during bonding as can be seen in Fig. 4. The γ' layer formed both in samples welded without any interlayer and in those with a nickel interlayer at the mating surfaces. It permanently pinned the weld interface to its original, planar configuration

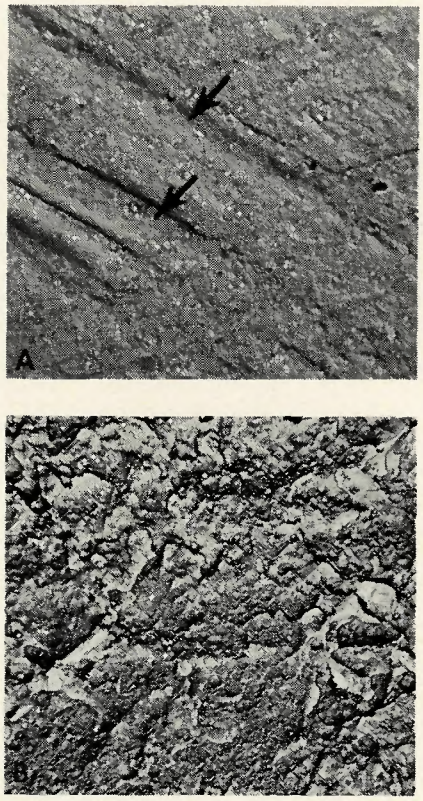

Fig. 3—Replica fractographs of broken joint surfaces in samples of wrought Udimet 700 bonded for 5 minutes at 2140F and 1 ksi. Sample (A) bonded without an interlayer. Arrows indicate unbonded grooves produced during surface preparation. Sample (B) bonded with a nickel interlayer. Mag: 5000X

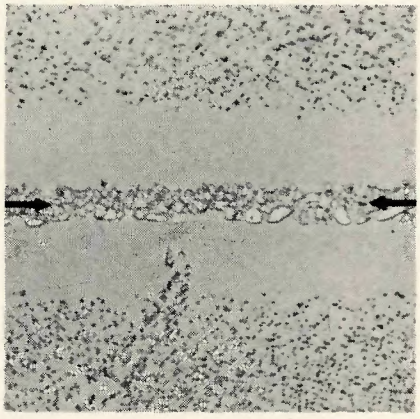

Fig. 4—As-welded microstructure of wrought Udimet 700 specimen bonded below the γ-γ' solvus at 2000F/2 hours/3 ksi with a 5μ nickel interlayer. Note the γ' layer formed at the planar joint interface and the precipitate-depleted region adjacent to it. Mag: 1000X

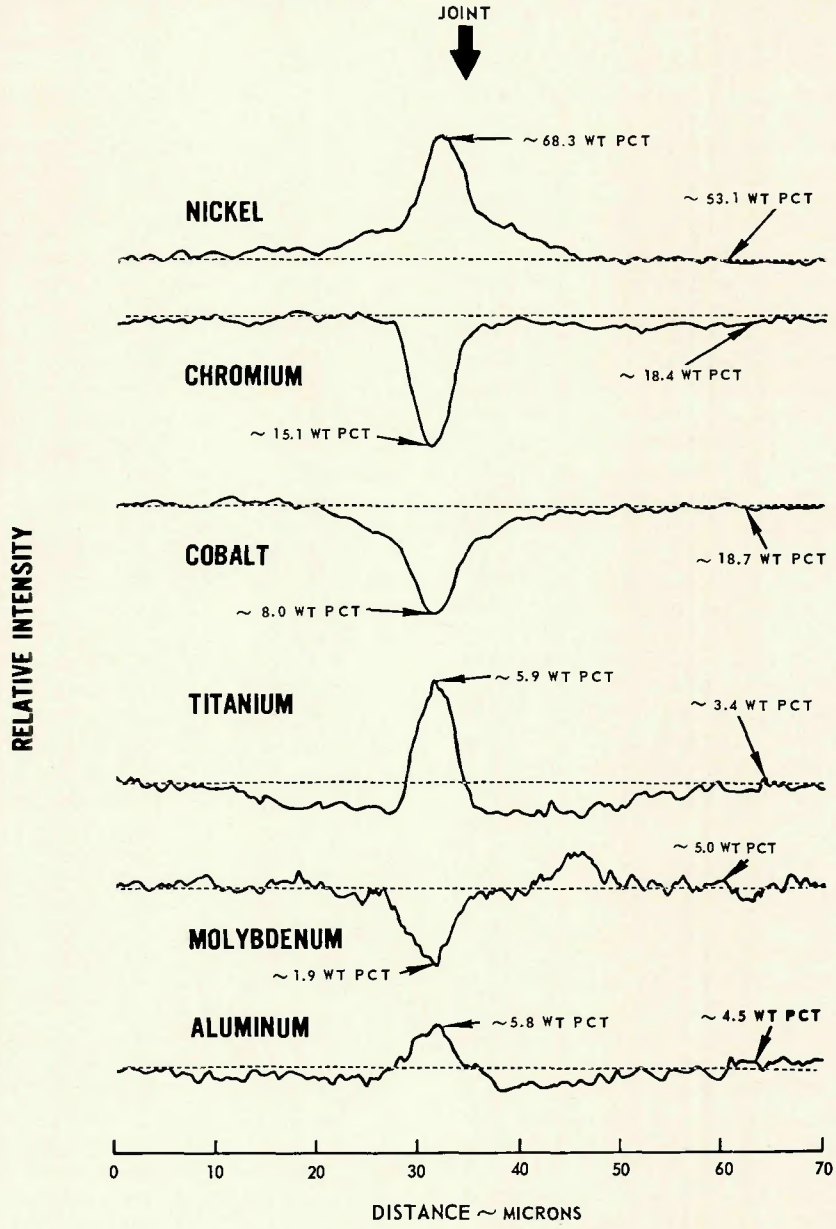

Fig. 5—Composition gradients across the joint area of wrought Udimet 700 sample bonded with a 5μ nickel interlayer at 2000F (sample in Fig. 4)

and resulted in joint brittleness. The formation of the γ' layer was unexpected as it was felt that in the nickel interlayer the concentration of γ' forming elements would be low.

However, microprobe analysis, as seen in Fig. 5, showed the contrary to be true. Due to preferential diffusion of titanium and aluminum into the nickel interlayers, there was actually an enrichment of these elements at the joint. This led to the massive γ' formation observed. Consequently, welding temperatures were kept above the γ-γ' solvus (approximately 2050 to 2100F) for subsequent work.

Since the wrought alloy is commonly solution heat treated at 2140F for 4 hours, this temperature and time (2140F/4 hours) were selected for most of the bonding experiments. A 1 ksi compressive stress was used for welding many of the wrought alloy specimens as this pressure produced satisfactory joints yet limited macroscopic deformation of the samples to <1% increase in cross-sectional area. Cast alloy samples were also welded at 2140F/4 hours/1 ksi.

However, it was later found that considerable improvement could be achieved in the quality of diffusion welds in the cast alloy if the amount of deformation during bonding was increased to 5 to 10%. Thus somewhat higher temperatures and pressures were used for some joints in the cast alloy. (This is described in detail subsequently.)

To obtain satisfactory boundary migration in Udimet 700, the weld specimens had to be maintained under a compressive stress for the duration (usually 4 hours) of the bonding cycle. The interface migration which occurred during the diffusion welding was apparently caused in large part by stress-induced grain boundary movement. This is contrary to the behavior of titanium diffusion welds where stress-free annealing of joints containing porosity causes substantial interfacial migration and pore elimination.[2]

A bonding cycle of 4 hours at 2140F and a 1 ksi compressive stress produced relatively complete joint homogenization and good interfacial boundary migration in samples of the wrought alloy. A micrograph of a diffusion weld produced in nickel plated wrought Udimet 700 with these bonding parameters is shown in Fig. 6. This sample was given a full aging heat treatment after welding. It can be seen that the interfacial grain boundary (Fig. 6) has migrated extensively from the original joint plane. The only remaining evidence of the original joint location in the wrought specimen is the darkly etched band.

In specimens of the cast alloy bonded with these parameters, the original joint area was delineated by some microstructural heterogeneity in the distribution and appearance of the γ' precipitate. Both the darkly etching band and the microstructural heterogeneity in the samples welded with a nickel interlayer were associated with residual chemical heterogeneity at the location of the original bond interface.

The data in Fig. 7 from electron-probe microanalysis across the wrought alloy joint illustrate the composition gradients which were present. The nickel concentrations are slightly higher at the joint than in the base metal while the molybdenum, chromium and cobalt concentrations at the joint are somewhat lower than in the parent metal. The aluminum and titanium concentrations in the wrought sample remain higher at the joint than in the base metal.

Although nickel interlayers permitted relatively high quality diffusion welds to be made in both cast and wrought samples, it was recognized that improvements were possible in several areas. Mechanical property studies indicated that these heterogeneous regions were sometimes preferential sites for transgranular cracking in elevated temperature tests. It is conceivable that some situations might also warrant diffusion welding at temperatures deliberately below the γ-γ' solvus of the base metal. Consequently, an attempt was made to find an easily applied interlayer which would reduce the tendency for interfacial γ' formation during diffusion welding, permit normal homogeneous precipitation of the strengthening γ' at the interlayer location during post-weld heat treatment and improve the chem-

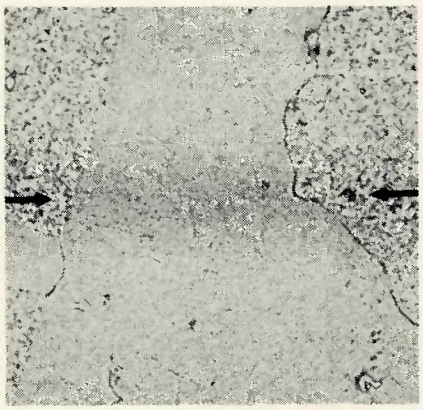

Fig. 6—Microstructure of a diffusion weld in wrought Udimet 700 which was made at 2140F/4 hours/1 ksi with a 5μ nickel interlayer. Sample was given full post-weld heat treatment. Mag: 250X

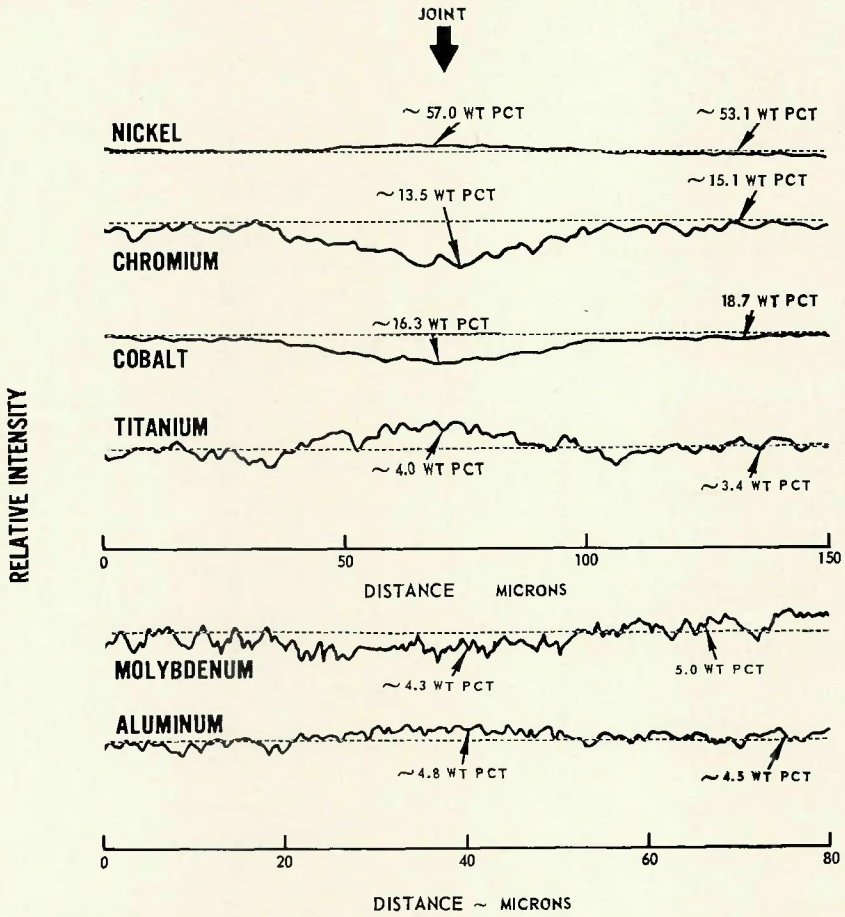

Fig. 7—Composition gradients across joint area of wrought alloy sample bonded with a 5μ nickel interlayer (sample in Fig. 6)

ical homogeneity of such elements as cobalt and chromium at the joint.

Cobalt strongly influences the γ' solvus temperature in nickel base superalloys.[6] Increasing cobalt contents up to ≈20% progressively raise the γ' solvus; however, at >20%, increases in cobalt substantially decrease the γ' solvus. Thus the use of an interlayer containing substantial cobalt could locally depress the γ' solution temperature and prevent interfacial γ' precipitation to much lower welding temperatures than with a nickel interlayer.

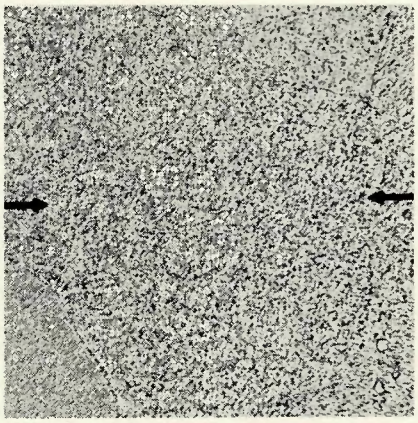

Fig. 8—Microstructural appearance of wrought Udimet 700 diffusion welded at 2140F/4 hours/1 ksi with a 5μ Ni-35%Co interlayer. Sample was fully heat treated after bonding. Mag: 250X

To test this hypothesis, samples were electroplated with 5 microns (total thickness) of cobalt and diffusion welded at 2140F and at 2000F. It was found that the cobalt interlayer successfully prevented interfacial γ' layer formation during bonding at 2000F. However, the cobalt concentrations remained high enough (even after postweld solution heat treatment at 2140F) to prevent substantial γ' precipitation during aging after diffusion welding..

A series of diffusion welds were then made with Ni-Co alloy interlayers to find if a balance of nickel and cobalt would suppress γ' layer formation during bonding but allow normal precipitation of this phase subsequently. Specimens were electroplated with 5 microns total thickness of various Ni-Co compositions (deposited as homogeneous alloys). The samples were then diffusion welded at either 2140F/4 hours/1000 psi or 2000F/2 hours/3000 psi.

The result of these tests demonstrated that the optimum composition range was Ni-25 to 35%Co for bonding Udimet 700. Interlayers within this composition range successfully prevented γ' layer formation at the joint interface during bonding both at 2000F and 2140F without upsetting normal γ' precipitation at this location during postweld aging.

A photomicrograph of a wrought specimen diffusion welded with a Ni-35%Co interlayer is shown in Fig. 8. The use of the Ni-Co interlayer resulted in much more chemical and microstructural homogeneity at the original joint location in both wrought and cast specimens than was achieved with a pure nickel interlayer. Note that the original interface location in the specimen in Fig. 8 is indistinguishable from the base metal and exhibits no evidence of either the darkly etched band (Fig. 6) or the microstructural heterogeneity observed in the samples bonded with a nickel interlayer.

The improved chemical homogeneity of this diffusion weld can be seen in the data (Fig. 9) from an electron-probe microanalysis across the original interface location in this specimen. A U.S. Patent for this diffusion welding process incorporating the Ni-Co interlayer has recently been issued.[7]

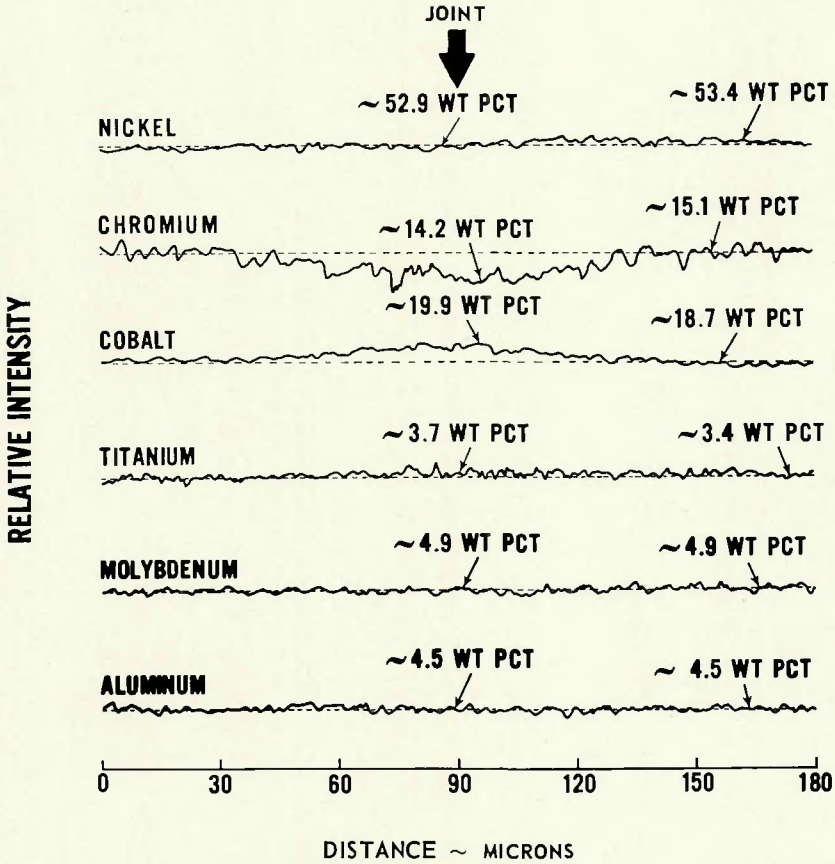

Fig. 9—Electron-probe microanalysis across diffusion weld in wrought Udimet 700 sample bonded with a 5μ Ni-35%Co interlayer (sample in Fig. 8)

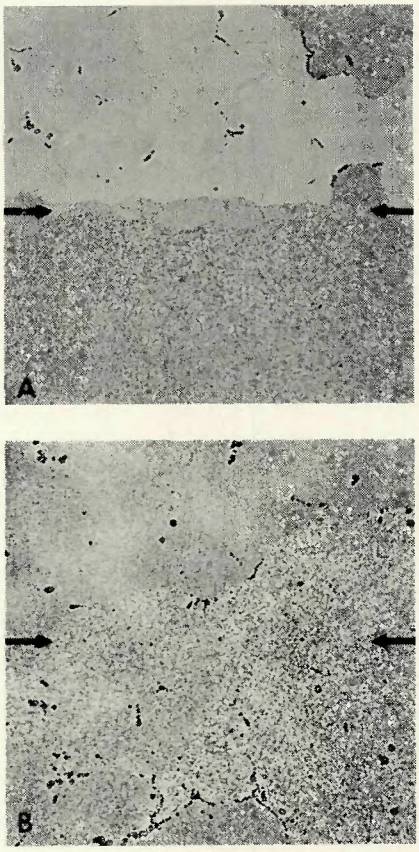

Fig. 10—Diffusion welds made in cast Udimet 700 with Ni-Co interlayer. Sample (A) bonded at 2140F/4 hours/1 ksi (<1% deformation). Note planar nature of interfacial area. Sample (B) bonded with improved parameters of 2175F/4 hours/1.2 ksi (≈7% deformation). Note increased amount of interfacial migration. Mag: 75X

Table 4—Tensile and Creep-Rupture Properties of Diffusion Welded Wrought Udimet 700

Tensile Tests

Sample[a]	Interlayer[b]	Test temp, °F	0.2% Y.S. ksi	U.T.S. ksi	Elong., %	R.A., %	Failure location
1	None		No test: failed at joint during machining				
2	None	1400	113	117	0.3	1.0	Joint
3	Nickel	Room temp	124	167	16.9	16.0	Base metal
4	Nickel	1400	114	137	11.1	11.1	Joint vicinity
Base metal requirements	—	Room temp	128	175	12	13	—
	—	1400	110	143	12	13	—

Creep-Rupture Tests at 1400F and 85 ksi

Sample[a]	Interlayer[b]	Rupture life, hr	Elong., %	Red. of Area, %	Failure location	Estimated joint efficiency[c]
5	None		No test: failed at joint during machining			
6	None	6.2	1.9	2.6	Joint	86%
7	Nickel	14.1	2.8	2.2	Joint vicinity	92%
8	Nickel	9.8	2.4	3.0	Joint vicinity	88%
Base metal requirements	—	40	5	—	—	—

[a] All specimens were diffusion welded at 2140F/3 hr/1 ksi; then heat treated as follows: 2140F/1 hr + 1975F/4 hr + 1550F/4 hr + 1400F/16 hr
[b] All interlayers were deposited by electroplating to 5 microns total thickness
[c] Determined from estimate of reduction of stress necessary to increase failure time to sample in question to specification requirements

Although the use of Ni-Co interlayers and the previously established joining parameters (2140F/4 hours/1 ksi) produced satisfactory diffusion welds in the wrought alloy, the joints formed in cast Udimet 700 were still more planar than desired (Fig. 10a). Testing confirmed that while cast specimens (bonded with these parameters) had properties approaching those of the base metal at 1400F, their creep properties at higher temperatures were deficient because of the planarity of the interfacial area.

Grain-boundary motion was more sluggish in the cast alloy than in the wrought material. Therefore, it was necessary to increase the welding temperature and pressure to cause migration of the interfacial boundary to a more satisfactory configuration in cast alloy diffusion welds.

Raising the welding temperature to 2175F and the compressive pressure to 1.2 ksi while using a 4 hour bonding cycle substantially increased interfacial migration as can be seen in Fig. 10b. These improved welding parameters (2175F/4 hours/1.2 ksi) increased the amount of sample deformation to 5 to 10% increase in cross-sectional area (compared to <1% during bonding at 2140F/4 hours/1 ksi). These mechanical property tests described in the next section demonstrate that increased sample deformation (up to 10%) during welding correlates with increased joint quality and creep life.

Mechanical Properties

A select group of mechanical properties tests were performed concurrently with the metallographic investigation. Tensile and creep-rupture tests were conducted to assist in defining the quality of diffusion welds made under various conditions. The initial mechanical tests were performed primarily on samples of the wrought alloy which were joined using a nickel interlayer and the preliminary bonding parameters (2140F/3 hours/1 ksi). The effect of improved welding techniques and Ni-Co interlayers were evaluated later with cast specimens. As the testing accompanied the process development, verification of this improvement on wrought alloy samples was not obtained at the time of this writing. However, the improvement in properties found in cast alloy diffusion bonds made with Ni-Co interlayers would be expected to be found in the wrought alloy.

Results of tensile and creep-rupture tests of diffusion-welded wrought Udimet 700 are listed in Table 4. These data illustrate the extremely poor properties of diffusion welds made without an interlayer. Of the two tensile and two creep-rupture specimens made without an interlayer, two broke during machining while the tensile specimen (sample 2 in Table 4) which was tested failed at substandard strength and with little ductility. In contrast, a room-temperature and a 1400F tensile sample diffusion welded with a nickel interlayer (samples 3 and 4 in Table 4) exhibited properties nearly equivalent to those of the base metal.

Because experience has shown that tensile tests of diffusion welds do not always accurately assess joint quality, creep-rupture testing was utilized to provide a more severe and discriminatory evaluation of mechanical properties. Creep-rupture tests were conducted on several wrought and cast diffusion welded specimens bonded under various conditions. The first series of creep-rupture tests were performed at 1400F and 85 ksi stress.

Results of these tests on the wrought alloy are listed in Table 4. The creep-rupture data for the wrought specimens illustrate the improvement gained through use of an interlayer. While the wrought specimens diffusion welded with a 5-micron nickel interlayer (samples 7 and 8) did not achieve base metal creep-rupture properties, the estimated joint efficiencies exhibited by these two specimens were ≈90% of base metal requirements. Metallographic examination of samples 7 and 8 revealed that fracture occurred both intergranularly and transgranularly in the vicinity of the joint. The transgranular fracture took place through the heterogeneous, darkly etched band observed in diffusion welds made with a pure nickel interlayer.

Diffusion welds made in cast Udimet 700 were also creep-rupture tested at 1400F. Specimens bonded with both nickel and Ni-Co electroplated mating surfaces were tested. In all cases, in the 1400F tests, the diffusion welds in the cast alloy had joint efficiencies which were estimated at ≥90% of base-metal requirements. The failure times and joint efficiencies of these specimens are compared to the cast base metal stress-time curve in Fig. 11.

To verify that the diffusion welded cast specimens did have joint efficiencies ≥90%, a specimen was creep-rupture tested at 1400F and 75 ksi. This stress (75 ksi) is ≈ 88 %of the 85 ksi at which base metal specification life is 23 hours. This sample failed in 121.8 hours, thus exhibiting

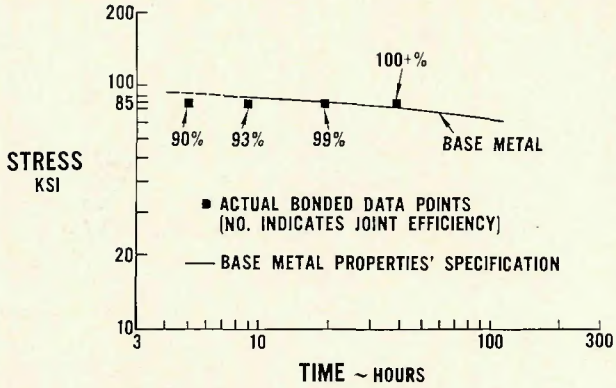

Fig. 11—Creep-rupture results obtained for diffusion welded cast Udimet 700 at 1400F and 85 ksi compared to base metal property requirements. Calculated joint efficiencies of each diffusion welded specimen are also indicated

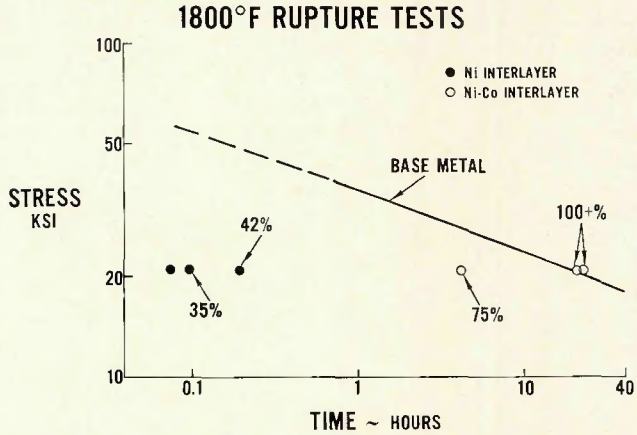

Fig. 12—Creep-rupture results obtained for diffusion welded cast Udimet 700 at 1800Fand 21 ksi compared to base metal property requirements. Calculated joint efficiencies of each specimen are indicated

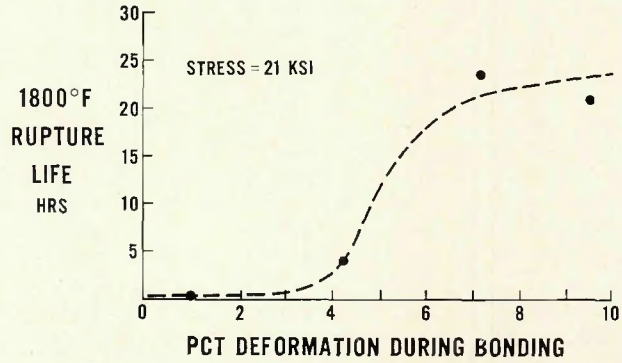

Fig. 13—Creep-rupture life at 1800F/21 ksi in diffusion welded cast Udimet 700 as a function of deformation during bonding

>>88% joint efficiency. In fact, the rupture life of this sample exceeded the 75 ksi base metal specification life of ≃ 100 hours.

In the cast Udimet 700 specimens creep-rupture tested at 1400F, fracture often occurred near the joint and frequently followed the interfacial grain boundary. The microstructural characteristics of these joints were not considered optimum because of limited interface migration (Fig. 10a) even though the specimens exhibited creep-rupture lives at 1400F which approached those of the base metal. Subsequent creep-rupture tests at 1800F and 21 ksi on three cast specimens bonded with a nickel interlayer and the initial parameters (2140F/3 hours/1 ksi) illustrated the deficiency of these semi-planar joints in higher temperature creep. These three specimens made with nickel interlayers failed in short times with negligible ductility and exhibited only 30 to 40% joint efficiencies (Fig. 12).

By increasing bonding temperature, pressure and time (to 2175F/4 hours/1.2 ksi) and using the Ni-Co interlayer, joint quality and the amount of interfacial migration were substantially improved as was previously discussed. Three of the improved joints were creep-rupture tested at 1800F because this temperature provided a more rigorous test than that at 1400F.

Test results are compared in Fig. 12 with the rupture lives of specimens welded with nickel interlayers, and the improvement is apparent. One sample possessed ≃ 75% joint efficiency while the other two specimens had >100% efficiencies, i.e., their rupture lives exceeded the 20-hour base metal specification requirement. Fracture in these specimens was intergranular and probably occurred near the original joint (it was difficult to metallographically ascertain the joint location in these samples). Ductility was improved and the fracture surfaces were considerably more irregular than the semi-planar fractures of the previous samples.

Analysis of the 1800F creep-rupture tests indicated that joint quality and rupture life were related to the amount of deformation induced during welding. Figure 13 illustrates the improvement in rupture life as sample deformation at the joint increased. From these data it appears that plastic deformation during bonding of 5 to 10% (increase in cross-sectional area) produce the best creep-rupture properties in diffusion welded cast Udimet 700.

Discussion

Results of this investigation have shown that strong diffusion welds can be obtained in high strength wrought and cast Udimet 700. Elevated temperature tensile and creep-rupture tests demonstrated that joint efficiencies of ≧ 90% were achieved when proper bonding techniques were employed. It was found that void-free joints with complete metallurgical bonding were easily obtained using a variety of welding conditions. Joint properties, however, strongly depend on other metallurgical qualities of the diffusion weld. To produce high strength diffusion welds, it was necessary to not only produce complete, void-free bonds but also to cause considerable movement of the original joint boundary and homogenization of the interfacial area. While these latter factors were more difficult to obtain, techniques were developed which resulted in strong diffusion welds with the desired characteristics.

It was found that an interlayer material was necessary between the mating base metal surfaces to prevent the formation of Ti(C,N) and $NiTiO_3$ interfacial precipitates during bonding. In samples made without an interlayer, these precipitates pinned the interfacial grain boundary to the original joint plane which resulted in very brittle joints. The composition, thickness and method of application of the required interlayer had to be carefully controlled to maximize joint quality. The use of 5 microns of pure nickel as the interlayer resulted in

joints with a composition (at the location of the original pure nickel) which was close to that of the Udimet 700 base metal (Fig. 7). However, the slight compositional difference which did exist altered the γ' morphology and distribution especially in the cast material.

Metallographic examination of the creep-rupture samples showed that this heterogeneous joint area was a location for transgranular fracture in both wrought and cast specimens. The use of a Ni-Co interlayer containing 25 to 35% cobalt improved the chemical homogeneity of the diffusion bonded joints and substantially increased the microstructural homogeneity of the joints in cast Udimet 700 compared to that obtained with a pure nickel interlayer. This improvement could be due to several factors including a possible increase in the effective diffusivity of alloying elements because of the cobalt addition. In addition, the Ni-Co interlayer removed the danger of an embrittling γ' layer forming at the joint and restricting its mobility should the diffusion welding temperature drop below the γ-γ' solvus of the base metal during the bonding operation. The Ni-Co interlayer also permits the option of welding at temperatures below the solvus should this be desirable.

The manner in which stress was applied during welding strongly influenced the amount of migration of the joint interfacial boundary and the resultant joint quality. Raising the bonding stress increased the interfacial boundary migration and improved the strength of the weld. These improvements correlated with the observed greater macroscopic deformation at the bond and the best weld quality was achieved at 5 to 10% plastic deformation (Fig. 12). However, the greatest boundary movement occurred when a constant compressive stress was applied to the mating surfaces throughout the entire welding cycle. A few samples were welded at a high initial stress which introduced $\simeq$ 5 to 10% deformation at the joint during the first 15 minutes of the bond cycle; the stress was then removed for the remainder of the 4-hour bond cycle. After welding, these specimens all contained semi-planar, continuous interfacial boundaries which exhibited limited migration. Thus it is necessary to maintain the bonding stress until sufficient boundary movement has occurred if high quality welds are to be achieved in this type of alloy.

The results of this study have demonstrated that it is feasible to diffusion weld high strength nickel base superalloys. While this investigation has dealt with only one alloy, Udimet 700, continuing studies indicate that it is equally feasible to achieve high quality diffusion welds in other high strength wrought and cast members of the superalloy family. The basic welding techniques discussed in the report are applicable to other alloys, although it is necessary to carefully select bonding temperatures, pressures and times to suit the characteristics of the alloy in question.

Conclusions

1. High quality diffusion welds can be successfully produced in both wrought and cast Udimet 700 by using a thin electroplated Ni-Co interlayer at the mating surfaces and carefully controlling bonding parameters.

2. In the limited tests conducted, the mechanical properties of diffusion welds in both the wrought and cast alloys approached those of the parent metal. Joint efficiencies $\geqslant$90% were obtained in room temperature and 1400F tensile tests on wrought samples. Likewise, $\geqslant$90% joint efficiencies were achieved in 1400F and 1800F creep-rupture tests on cast alloy diffusion welds.

3. The quality of joint obtained depends on the amount of interfacial grain boundary migration and the degree of chemical and microstructural homogenization achieved at the original interlayer location. The use of an electroplated Ni-Co interlayer improved joint homogeneity and reduced the sensitivity of joint quality to slight variations in processing parameters.

4. Results of this study indicate that diffusion welding is a promising method for joining complex components made out of high strength nickel base superalloys.

Acknowledgments

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Correction

The December, 1971 Welding Research Supplement article by Isasi, Bates and Heuschkel (p. 493-s) entitled "Cracking in Welded Steam Pipe" contained two minor errors.

The subtitle synopsis should read as follows: Cracking tends to occur in a new grade of austenitic stainless (Fe-Ni-Cr-Mn-Mo-V-N-B) when the carbon content exceeds 0.030% and when pipe having noncircular components is present in the system.

The piping discussed in the paper had a nominal wall thickness of 3/8 in. Since all micrographs and macrographs were reduced in reproduction, a misleading conclusion could possibly be drawn on this point.