Gas Tungsten Arc Welding of a Powder Metallurgy Aluminum Alloy

Fusion welds made by GTAW in powder metallurgy Al-10Fe-5Ce base metal prove inadequate for structural applications

BY G. E. METZGER

ABSTRACT. A thin sheet of the powder metallurgy aluminum alloy Al-10Fe-5Ce (wt-%) was fusion welded with the gas tungsten arc welding process. The effects of preweld vacuum heat treatment of the base metal, filler metal, and the type of welding current were investigated.

The gross porosity of welds made with untreated base metal was virtually eliminated by the use of a combination of preweld vacuum heat treatment and a direct current electrode negative (DCEN) welding arc with helium shielding gas. The required heat treatment is estimated to be 20 h at 750°F (399°C), which results in a reduction in the base metal tensile strength from about 65 to about 60 ksi (448 to 414 MPa).

However, the presence of brittle, closely spaced intermetallic compounds near the weld interface causes low ductility and low strength. The tensile elongation at room temperature of longitudinal welds was about 1%, and the tensile strength joint efficiency of transverse welds was 55% when tested at room temperature, and 81% at a testing temperature of 600°F (316°C).

Introduction

Recently developed aluminum alloys, produced by powder metallurgy methods, have higher strength at elevated temperature than aluminum alloys produced by conventional means. The principal cause of the higher strength is the presence of finely dispersed intermetallic particles. Joining of these powder metallurgy alloys is one of the requirements for their efficient application in structures. A feasibility study of the gas tungsten arc welding of one of these alloys, a metastable Al-10Fe-5Ce (wt-%), is presented in this paper.

A persistent problem with fusion welds in aluminum base metal made by the compaction of powder has been the formation of excessive gas porosity in the weld metal (Refs. 1–6) caused by the presence of hydrogen, primarily as hydrate or hydroxide, in the base metal. The hydrogen content may be reduced to an acceptable level by heat treatment of the base metal in vacuum at high temperature (Ref. 4). However, this heat treatment is limited by the reduction of base metal strength, as the time and temperature increase. An unacceptable hydrogen level can be avoided by the use of an inert atmosphere in the manufacture of the powder from which the base metal is made (Refs. 5, 6).

Fusion welding of the metastable powder metallurgy aluminum alloys will often result in the conversion of the fine dispersoids into coarse intermetallic compounds in the fusion zone of the weld, to the detriment of the mechanical properties of the weld (Refs. 7, 8). However, this behavior may be sufficiently suppressed, for very thin base metal, by electron beam welding with low energy input to produce welds with good strength and ductility (Ref. 8).

KEY WORDS

GTAW
Al-10Fe-5Ce
Preweld Vacuum Heat
Filler Metal
Welding Current
Tensile Strength
Powder Metallurgy
Aluminum Alloys
Porosity

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Materials

The base metal used in this investigation was the aluminum alloy, Al-10Fe-5Ce (wt-%), consolidated from powder by extrusion to form a bar of rectangular cross-section, ¾ X 4 in. (19 X 102 mm). All welding and tensile testing was done with strips, 1/16 X ¾ X 4 in., made by mechanical cutting perpendicular to the longitudinal axis of the bar, followed by machining to the 1/16-in. (1.6-mm) thickness.

Vacuum fusion analysis yielded a hydrogen content of 2 ppm, which was reduced to 1 ppm by treatment in a vacuum at 750°F for 100 h. Filler metals included Classifications ER4043 (5% Si), ER4047 (12% Si), and ER5356 (5% Mg) of the American Welding Society specification, AWS A5.10-80.

Base Metal Tensile Properties

Tension specimens, as shown in Fig. 1, were tested. There was a strong tendency for the specimens to fracture near the end of the 1-in. (25-mm) gauge length, thus making it impossible to determine the elongation, except by the use of an unconventional procedure (Ref. 9). The entire length of the reduced section was subdivided at 0.1-in. (2.5-mm) intervals by marking with a scribe before tensile testing. Figure 2A illustrates a tension specimen with the fracture near the end of the 1-in. gauge length. Dimension a represents the extended gauge length that would be used for calculation of the elongation, if the specimen had fractured near the center of the gauge length. Dimension *b* of Fig. 2B represents the extended gauge length that is used to compensate for fracture near the end of the gauge length, and is composed of three elements. These are: dimension *c* of five scribe marks, which is one-half of the extended gauge length on the longer end of Fig. 2A; dimension *d* of two scribe marks, which is the dis-

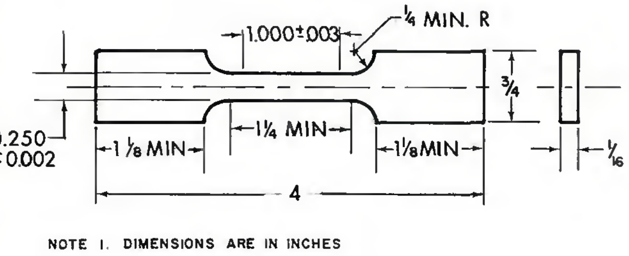


Fig. 1 — Tension specimen.

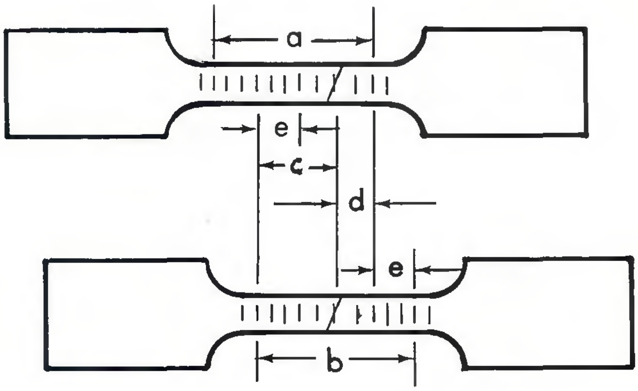


Fig. 2 — Calculation of tensile elongation.

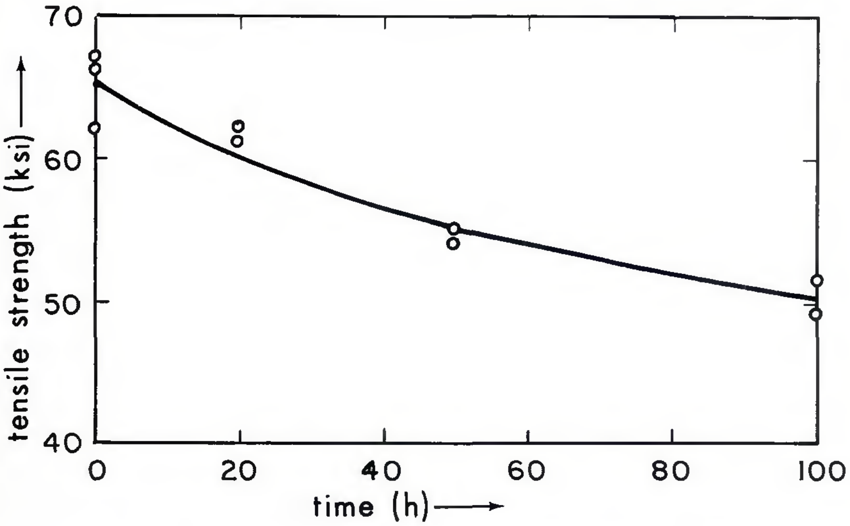


Fig. 3 — Base metal tensile strength vs. heat treatment time at 750°F.

tance between the scribe mark nearest the fracture and the end of the gauge length on the shorter end of Fig. 2A; and dimension *e* of three scribe marks, to complete the required ten scribe marks of dimension *b* of Fig. 2B.

The results of base metal tension tests are presented in Table 1. The holding time before testing was not recorded. The results are in good agreement with those given by Lockheed (Ref. 10) for the same alloy, except that a tensile strength of 32 ksi at 600°F (221 MPa at 316°C) was reported by Lockheed. Since that value is consistent with the results of extensive testing of a series of Al-Fe-Ce alloys by Lockheed, it is considered that 32 ksi is more accurate than the 50–52 ksi (345–359 MPa) shown in Table 1.

One tension specimen, No. 16, tested after heat treatment at 750°F for 100 h in air, had a tensile strength very near that of Nos. 13, 14 and 15, which were heat treated at the same time and temperature, but in vacuum. This indicates that, at these heat treatment conditions, air and vacuum have about the same effect on tensile strength.

Figure 3 shows, in graphical form, the effect of heat treatment at 750°F in vacuum on the tensile strength.

Table 1—Results of Base Metal Tension Tests

Tension Specimen No.	Heat Treatment Temp. °F	Heat Treatment Time h	Heat Treatment Atmosphere	Test Temp. °F	Yield Strength, ksi	Tensile Strength, ksi	Elong., %(1)
1				RT		67	
2				RT		62	
3				RT	51	66	7.9
4				RT	53	66	7.5
5				400		57	
6				400		57	
7				600		52(2)	
8				600		50(2)	
9	750	20	vacuum	RT		62	
10	750	20	vacuum	RT		61	
11	750	50	vacuum	RT		54	
12	750	50	vacuum	RT		55	
13	750	100	vacuum	RT		51	
14	750	100	vacuum	RT		51	
15	750	100	vacuum	RT	35	49	
16	750	100	air	RT	37	48	10

(1) Elongation in 1-in. gauge length.
(2) See text.

Welding

Mechanized welding with the gas tungsten arc welding (GTAW) process was used. The specimens were held in a clamping fixture with shielding gas supplied both to the welding torch and to the weld root. The only preweld surface preparation of the base metal strips was degreasing in acetone.

Welds Made with Alternating Current (AC)

Autogeneous Welds

The first group of welds was made to obtain preliminary information on the effect of preweld heat treatment of the

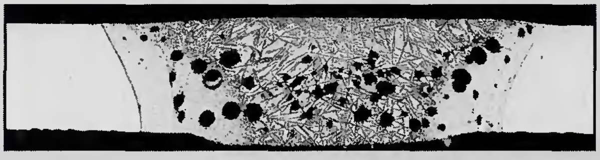


Fig. 4 — Autogeneous weld made with no preweld heat treatment, AC (no etchant, 11X).

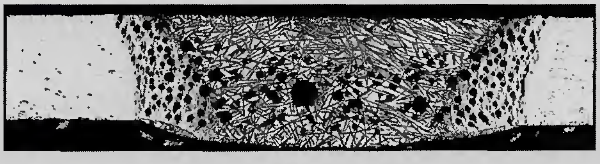


Fig. 5 — Autogeneous weld made with preweld heat treatment at 750°F for 1000 h in air, AC (no etchant, 11X).

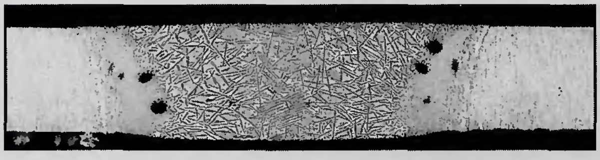


Fig. 6 — Autogeneous weld made with preweld heat treatment at 750°F for 10 h in vacuum, AC (no etchant, 11X).

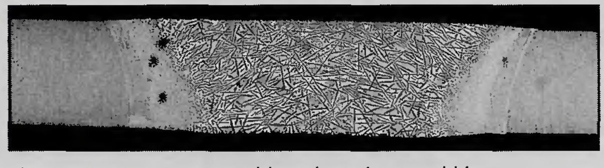


Fig. 7 — Autogeneous weld made with preweld heat treatment at 750°F for 100 h in vacuum, AC (no etchant, 11X).

base metal on weld porosity. Welds were made with no filler metal, with argon shielding gas, and in the flat position along the longitudinal axis of the base metal strip. Alternating current (AC) of 115–120 A as the welding current, 15–18 V as the arc voltage, and a welding speed of 5 in./min (2.1 mm/s) were used to obtain a depth of fusion through the strip thickness.

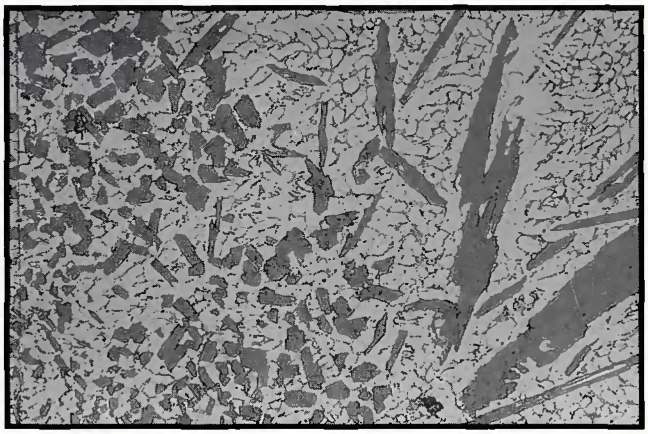
Preweld heat treatment of the base metal strips included exposure at temperatures of both 600° and 750°F for times of 10 and 100 h in vacuum and 10, 100 and 1000 h in air.

The welded strips were radiographed, and then cut perpendicular to the longitudinal axis into four 1-in. (25-mm) segments. Each of the three resulting cross-sections was examined as a metallographic mount. Radiographs made with a variety of exposure conditions did not prove to be useful as a means of examination for weld porosity.

The excessive porosity of a typical cross-section of a weld made with no preweld heat treatment is shown in Fig. 4. A typical cross-section of a weld made after a preweld heat treatment at 750°F for 1000 h in air is presented in Fig. 5. Although heat treatment in air caused a reduction in the pore size, there appeared to be little change in the porosity volume. Welds made with base metal that had been heat treated in air at 750°F for less than 1000 h, as well as at 600°F, yielded about the same porosity results as shown in Figs. 4 and 5.

A preweld heat treatment in vacuum at 750°F for 100 h resulted in welds with a marked decrease in the porosity volume, when compared to no preweld heat treatment, but 100 h at 600°F resulted in little or no improvement.

The encouraging results with base



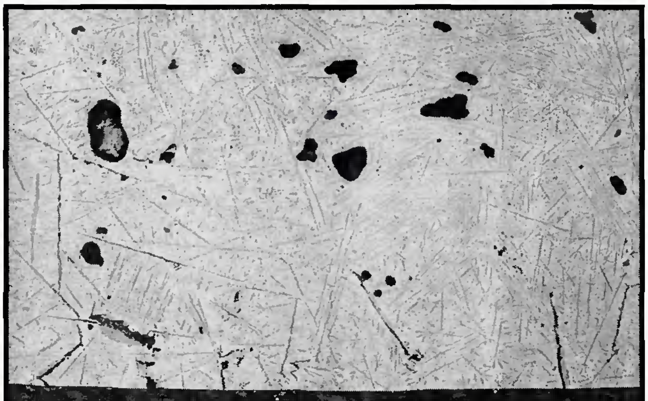


Fig. 8 — Top left: Microstructure of AC autogeneous weld at the weld toe area.

Fig. 9 — Above: Transition from blocky to acicular form of $FeAl_3$.

Fig. 10 — Left: Microcracks extending from root surface along acicular $FeAl_3$ intermetallics.

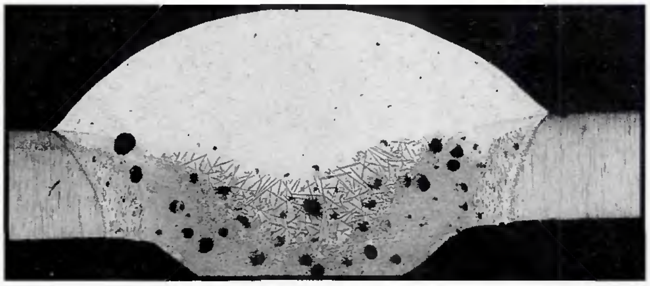


Fig. 11 — Weld made with no preweld heat treatment, AC, ER4043 (no etchant, 11X).

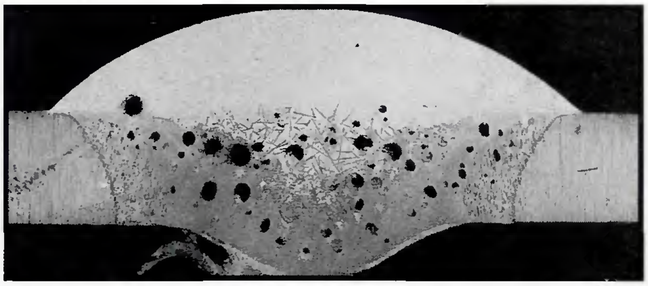


Fig. 12 — Weld made with no preweld heat treatment, AC, ER5356 (no etchant, 11X).

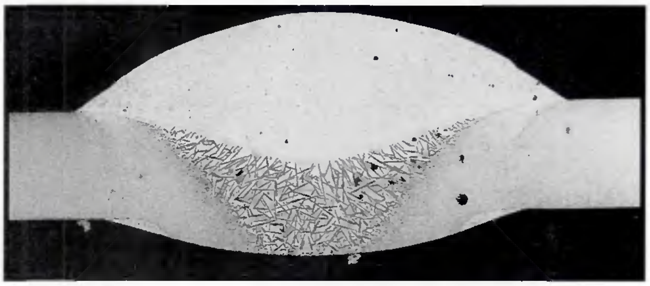


Fig. 13 — Weld made with preweld heat treatment at 750°F for 100 h in vacuum, AC, ER4043 (no etchant, 11X).

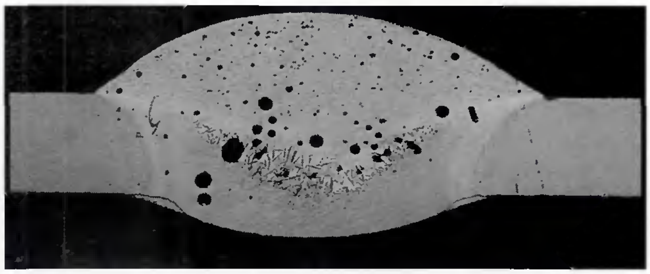


Fig. 14 — Weld made with preweld heat treatment at 750°F for 100 h in vacuum, AC, ER4047 (no etchant, 11X).

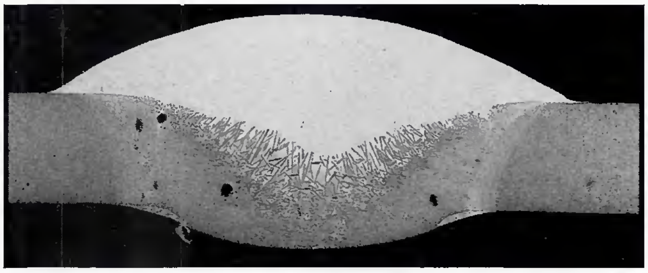


Fig. 15 — Weld made with preweld heat treatment at 750°F for 100 h in vacuum, AC, ER5356 (no etchant, 11X).

Table 2—Welding Conditions for Welds Made with AC and Filler Metal

Weld No.	HT Time,(1) h	Arc Filler Metal	Welding Voltage, V	Current, A
1	none	ER4043	18	125
2	none	ER4047	21	135
3	none	ER5356	15	165
4	10	ER4043	18	125
5	10	ER4047	21	135
6	10	ER5356	15	165
7	100	ER4043	18	125
8	100	ER4047	21	135
9	100	ER5356	15	165

(1) Preweld heat treatment at 750°F in vacuum.

metal heat treated at 750°F in vacuum led to a second group of welds with preweld heat treatment using those conditions at 10, 20, 50 and 100 h. The welding conditions were the same as for the first group. Photomicrographs of typical cross-sections, prepared as described for the first group of welds, illustrating the beneficial effect of vacuum heat treatment at 750°F, are presented in Fig. 6 for a 10 h heat treatment and in Fig. 7 for 100 h. Welds in base metal with the 20-h heat treatment were about the same as those for 10 h, and welds with 50 h were about midway between those for 10 h and 100 h. It is noted that most of the porosity in these welds is located in the fusion zone near the weld interface.

One of the AC autogeneous welds, made with base metal that had been exposed to vacuum for 1000 h at 450°F, was selected for detailed examination of the weld toe area. The microstructure of the entire weld was similar to the weld of Fig. 4. The microstructure of the weld interface, at the weld toe, is shown in Fig. 8. At higher magnification, the transition of the phase in the blocky gray form to the phase in the acicular gray form is shown in Fig. 9. Both of those phases were identified as the intermetallic compound $FeAl_3$ by electron probe analysis. The blocky form is believed to be a cross-section of the acicular form. The microstructure at the weld toe area of all welds made later in this investigation was similar to that of Fig. 8.

In the same weld, microcracks extending from the root surface along the intermetallic needles were present — Fig. 10.

Welds Made with Filler Metal

A third group of welds was made to determine the effect on weld porosity of a combination of preweld heat treatment at 750°F in vacuum and the use of filler metal. The welding conditions were the same as for the first group, except as shown in Table 2. Preparation of the welds for metallographic examination was the same as for the first group.

A representative selection of photomicrographs from these welds is presented in Figs. 11 through 15. Figs. 11 and 12 illustrate the typical excessive porosity of welds made with no preweld heat treatment of the base metal. However, the upper part of the weld, consisting mostly of filler metal, is almost free of porosity. The porosity in the weld made with ER4047 was about the same as those made with ER4043 and ER5356.

Figures 13 through 15 illustrate the typical low porosity in welds made with preweld heat treatment at 750°F for 100 h. Welds made with ER4043 and ER5356 had about the same amount of porosity as autogeneous welds (Fig. 7), and were much better than welds made with ER4047. Again, the porosity was confined mostly to the weld metal zone containing a high proportion of base metal.

There was no significant decrease in porosity for welds made with preweld heat treatment at 750°F for 10 h, as compared to no preweld heat treatment.

Welds Made with Direct Current Electrode Negative (DCEN)

Welds were made in the flat position at a travel speed of 5 in./min, and with filler metal. Direct current electrode negative (DCEN) with helium shielding gas was used to obtain a depth of fusion through the base metal thickness.

The first group of DCEN welds, along the longitudinal axis of the base metal strip, was made to determine the effect of preweld heat treatment and filler metal on weld porosity. The welding conditions for these longitudinal welds are given in Table 3 for weld numbers 1 through 6. The filler metals selected were ER4043 and ER5356, since welds made with these filler metals and AC contained less porosity than those made with ER4047.

A representative selection of photomicrographs of these welds is shown in Figs. 16–19. The use of direct current resulted in a remarkable decrease in weld porosity, as compared to alternating current, for welds made with no preweld heat treatment; and, an appreciable decrease for welds made with preweld vacuum heat treatment. The weld illustrated in Fig. 17 appears to indicate that weld porosity is virtually eliminated, when DCEN and ER5356 is used with a preweld heat treatment in vacuum at 750°F for 100 h.

Other significant differences are the much more thorough mixing of base metal and filler metal in the weld metal, and the greater uniformity, along the weld length, in the width and height of the root reinforcement in welds made with direct current. In addition, welds made with ER5356 filler metal had slightly less weld porosity than those made with ER4043, and welds with a preweld heat treatment of 100 h also had slightly less weld porosity than those made with no preweld heat treatment.

Fig. 16 — Weld made with no preweld heat treatment, DCEN, ER5356.(Alcoa etch: nitric, hydroflouric, chromic acids; 11X).

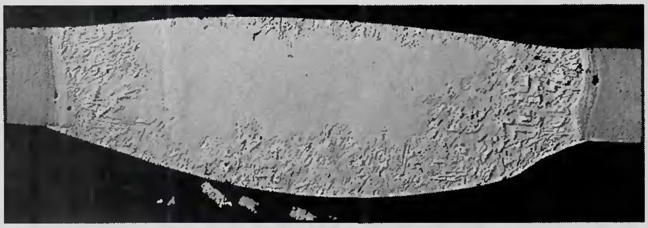


Fig. 17 — Weld made with preweld heat treatment at 750°F for 100 h in vacuum, DCEN, ER5356 (Alcoa etch: nitric, hydrofluoric, chromic acids; 11X).

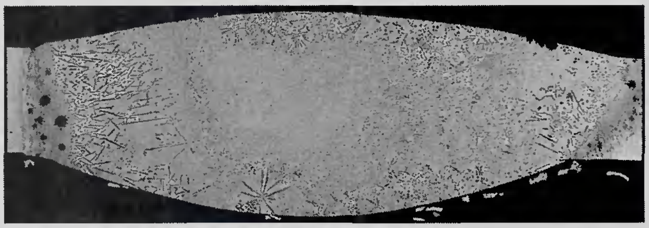


Fig. 18 — Weld made with no preweld heat treatment, DCEN, ER4043 (no etchant, 11X).

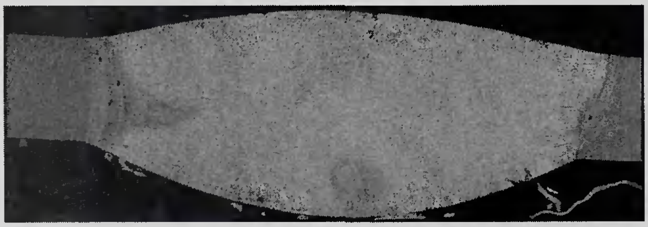


Fig. 19 — Weld made with preweld heat treatment at 750°F for 100 h in vacuum, DCEN, ER4043 (no etchant, 11X).

Weld Tensile Properties

Welds were made in the flat position at a travel speed of 5 in./min, and with ER5356 filler metal. Direct current electrode negative (DCEN) with helium shielding gas was used to obtain a depth of fusion through the base metal thickness. There was no preweld heat treatment.

Two welds were made along the longitudinal axis of the base metal strip, and six welds were made transverse to the longitudinal axis and at the center of the base metal strip. The welding conditions for the two longitudinal welds are given with weld numbers 7 and 8, and for the six transverse welds with weld numbers 9 through 14 in Table 3.

The transverse welds were made by clamping a group of abutting strips, with machined straight edges, in the welding fixture and welding the entire group in one pass. This resulted in holes at some of the abutting edges, but sufficient undamaged welded strips remained to serve as tension specimens. Weld numbers 9 through 11 were in one group and 12 through 14 were in a second.

Tension specimens, as shown in Fig. 1, were made from weld numbers 7 through 14. The weld reinforcement was removed by milling from the two specimens with longitudinal welds, but was left intact for the six specimens with transverse welds. The results of the tension tests are shown in Table 4, with the specimen numbers corresponding to the weld numbers of Table 3.

Fracture of the transverse weld tension specimens occurred near the weld interface, through the zone containing the weld porosity. An examination of the fracture surfaces with a binocular microscope revealed considerably more weld porosity (see Fig. 20 for a typical example), as would be expected, than was evident in a cross-section of a weld made with similar welding conditions (see Fig. 16).

An examination of the fracture surfaces of the specimens with longitudinal welds indicated the formation of massive and closely spaced acicular $FeAl_3$ in the weld metal immediately adjacent to the weld porosity zone, as illustrated in Fig. 21, and then sometimes a zone containing the same phase in rosette form, as illustrated in Fig. 22. The orientation of the weld in both figures is with the weld face at the top of the photographs. The primarily fine-grained central portion of the weld metal is located at the left in Fig. 21 and at the right in Fig. 22. At the lower right corner of Fig. 21, which is located at the edge of the specimen and also at the weld root, a few pores may be observed. The low tensile strength of the welds is considered to be normal, even when any effect of weld porosity is not considered. The gauge section of the specimens with longitudinal welds (without weld reinforcement) is composed almost entirely of weld metal, which is an alloy of the lower strength ER5356 and the base metal. At the edge of the fusion zone, near the weld interface, the unmixed zone results in the formation of closely spaced $FeAl_3$ intermetallics from the finely dispersed intermetallic particles that are primarily responsible for the high strength of the base metal. This results in low strength of the specimens with transverse welds (with weld reinforcement), although it is expected that the weld porosity also contributed to the low strength.

There are several indications of brittle behavior in the results of the weld tension tests. The addition of the high ductility ER5356 filler metal to the base metal should result in an increase of weld metal ductility; yet, just the opposite is true. The average ductility of the longitudinal weld tension specimens (numbers 7 and 8 of Table 4) is about one order of magnitude less than the average ductility of the base metal (numbers 3 and 4 of Table 1). This decrease in ductility is too great to be attributed to the presence of a few pores on the weld cross-section. The failure of number 10 upon clamping in the tensile test fixture is a good indication of poor weld ductility. The increased joint efficiency at a testing temperature of 600°F is consistent with the fact that the effect of brittle behavior is often alleviated with an increase in temperature. And finally, the needles found on the fracture surfaces of the longitudinal weld tension specimens have been identified as the intermetallic compound $FeAl_3$. The brittle behavior of the coarse, concentrated in-

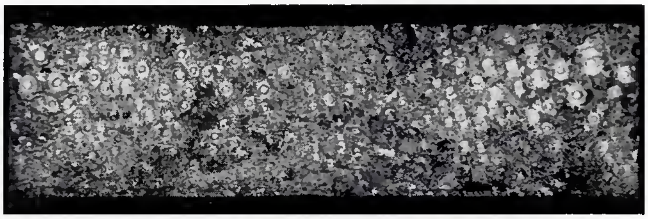


Fig. 20 — Weld porosity on fracture surface of tension specimen with transverse weld (16X).

Fig. 21 — Massive and closely spaced acicular $FeAl_3$ on fracture surface of tension specimen with longitudinal weld (N465) (32X).

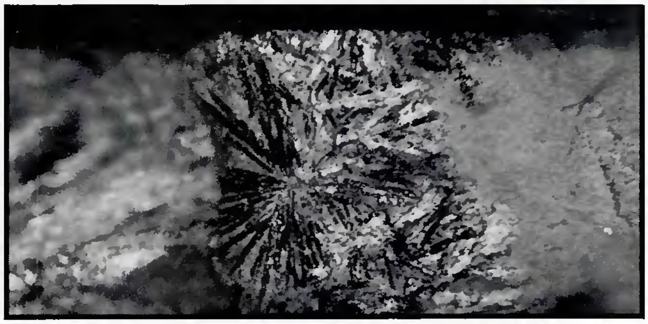


Fig. 22 — Acicular $FeAl_3$ in rosette form on fracture surface of tension specimen with longitudinal weld (N467) (32X).

Table 3—Welding Conditions for Welds Made with DCEN

Weld No.	HT Time,(1) h	Weld(2) Orientation	Filler Metal	Arc Voltage, V	Welding Current, A
1	none	coincident	ER4043	15	150
2	none	coincident	ER5356	14	185
3	10	coincident	ER4043	15	150
4	10	coincident	ER5356	14	185
5	100	coincident	ER4043	15	150
6	100	coincident	ER5356	14	185
7	none	coincident	ER5356	14	160
8	none	coincident	ER5356	13	150
9	none	transverse	ER5356	15	160
10	none	transverse	ER5356	15	160
11	none	transverse	ER5356	15	160
12	none	transverse	ER5356	18	140
13	none	transverse	ER5356	18	140
14	none	transverse	ER5356	18	140

(1) Preweld heat treatment at 750°F in vacuum.
(2) Relationship of weld axis to longitudinal axis of base metal strip.

termetallic compounds would cause premature failure and low ductility.

Discussion

The first welds made with the type of welding current (alternating current, AC) most commonly used for the gas tungsten arc welding of aluminum, and with no filler metal, exhibited gross porosity distributed throughout the weld metal, when there was no preweld heat treatment of the base metal and also when preweld heat treatment consisted of heating in air from a minimum exposure of 10 h at 600°F to a maximum of 1000 h at 750°F. Although preweld heat treatment in vacuum at 600°F did not appreciably affect the weld porosity, 750°F did drastically reduce the weld porosity, even at a time as short as 10 h. Rather than being scattered about throughout the weld metal, the porosity was confined to the zone immediately adjacent to the weld interface, where the short time in the liquid state, a temperature of the weld pool close to the melting point, and a rapid solidification rate resulted in the entrapment of gas bubbles.

There was a poor correlation between the apparent hydrogen content of the base metal and weld porosity. A vacuum heat treatment at 750°F for 100 h reduced the hydrogen content from 2 to 1 ppm; *i.e.*, the hydrogen was reduced by one half. However, the weld porosity reduction was an estimated two orders of magnitude. Apparently, there is something in the base metal, in addition to the hydrogen determined by vacuum fusion analysis, that causes weld porosity; or, the hydrogen analysis was not accurate.

The porosity in welds made after a preweld heat treatment of the base metal in vacuum at 750°F for 100 h is probably acceptable; however, the decrease in the tensile strength of the base metal from about 65 to 50 ksi caused by this treatment is not. Therefore, further weld experiments were made with three widely used commercial filler metals, again with AC, in an attempt to reduce weld porosity, and at the same time reduce the effect of the preweld heat treatment on the base metal strength.

Although the addition of filler metal did reduce weld porosity, it was still excessive, when either there was no preweld heat treatment or the sample was heat treated in vacuum at 750°F for 10 h. The shallow depth of fusion of the AC welding arc resulted in a weld bead that consisted primarily of filler metal at the weld-face side and primarily of base metal at the weld-root side. The greater agitation of the weld pool on the weld-face side permitted gas bubbles to escape, whereas the more stagnant weld metal on the weld-root side caused evolving gas to be entrapped, to form weld porosity. Welds made with a preweld heat treatment in vacuum at 750°F for 100 h and with ER4043 or ER5356 filler metals exhibited about the same weld porosity volume as autogeneous welds made with the same preweld heat treatment, but the pores were located throughout the portion of the weld metal consisting primarily of base metal, rather than concentrated near the weld interface.

With the objective of more uniform weld metal, the greater depth of fusion of the helium-shielded direct current electrode negative (DCEN) arc was used to make welds with ER4043 and ER5356 filler metals in base metal with no preweld heat treatment and with preweld heat treatment in vacuum. The objective was attained, but of much greater significance was the remarkable decrease in weld porosity. Welds with no preweld heat treatment had about the same amount of porosity as welds made with AC and a preweld heat treatment in vacuum at 750°F for 100 h, and porosity was virtually eliminated in welds with ER5356 filler metal and a preweld heat treatment in vacuum at 750°F for 100 h. The greater agitation through the entire depth of the weld metal, due to the greater depth of fusion of the helium-shielded DCEN arc, not only produced thorough mixing but also allowed all evolving gas to escape except a few bubbles entrapped near the weld interface, where molten time is short and solidification is rapid. An additional benefit of DCEN was better uniformity, along the weld length, in the width and height of the root reinforcement.

Transverse welds and longitudinal welds for tensile testing were made with DCEN, ER5356, and no preweld heat treatment. The fracture surfaces of tension specimens with a transverse weld, which failed near the weld interface through the zone containing weld porosity, revealed sufficient porosity to have some effect on the weld strength. Since either DCEN or preweld heat treatment in vacuum at 750°F for 100 hours resulted in welds with almost no porosity, it is probable that a combination of DCEN and preweld heat treatment for a time less than 100 h would also result in welds with no significant porosity. It is estimated that the necessary time would be on the order of 20 h, with a corresponding decrease in the base metal tensile strength from 65 to about 60 ksi.

However, a major deficiency in the welds, the presence of brittle intermetallic phases remains, in a zone of about 0.060-in. (1.5-mm) thickness near the weld interface, on the fusion side. This brittle zone causes low ductility and low strength. The tensile elongation at room temperature of longitudinal welds was about 1%, and the tensile strength joint efficiency of transverse welds was 55% when tested at room temperature, and 81% at a testing temperature of 600°F. If the thickness of a brittle layer can be limited to no more than a few mils, a weld will often exhibit good ductility. There is no possibility that welding at a greater travel speed to reduce the heat input would reduce the brittle zone to that thickness. It is also improbable that the lower heat input of plasma arc welding or electron beam welding would have sufficient effect on the thickness of the brittle zone to appreciably improve weld ductility, in thick base metal.

Conclusions

Fusion welds made by the GTAW process in powder metallurgy Al-10Fe-5Ce base metal appear to be of no use for structural applications. Although weld porosity can be virtually eliminated by a combination of DCEN welding and preweld vacuum heat treatment of the base metal, with only a

Table 4—Results of Weld Tension Tests

Tension Specimen No.	Weld[a] Orientation	Test Temp., °F	Yield Strength, ksi	Tensile Strength, ksi	Elong., %[2]	Joint Efficiency, %[3]
7	coincident	RT	18	20	0	
8	coincident	RT	19	20	1.6	
9	transverse	RT		36		55
10	transverse	RT		(4)		
11	transverse	400		33		52
12	transverse	400		26		
13	transverse	600	22			81
14	transverse	600	30			

(1) Relationship of weld axis to longitudinal axis of base metal strip.
(2) Elongation in 1-in. gauge length.
(3) A ratio, expressed in percent, of the average weld tensile strength to the average base metal tensile strength from Table 1, except that the base metal strength of 32 ksi at 600°F, determined by Lockheed, was used in the calculation.
(4) Specimen broke upon installation in tensile testing machine.

minor decrease in base metal tensile strength, the welds exhibit a lack of ductility due to brittle intermetallic phases formed near the weld interface in the fusion zone. It is probable that welds made in thick base metal by any fusion welding process would be useful only for sealing welds, or some other nonstructural application.

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