

To the Graduate Council:

I am submitting herewith a thesis written by Catherine Jessie Swindeman entitled "The Modification of Fusion Welding Techniques for Successful Joining of Silicon Carbide Particulate-Reinforced Aluminum Alloy Composites." I have examined the final copy of this thesis for form and content and recommend that it be accepted in partial fulfillment of the requirements for the degree of Master of Science, with a major in Metallurgical Engineering.



Carl D. Lundin, Major Professor

We have read this thesis and
recommend its acceptance:



Accepted for the Council:



Associate Vice Chancellor
and Dean of The Graduate School

THE MODIFICATION OF FUSION WELDING TECHNIQUES FOR SUCCESSFUL
JOINING OF SILICON CARBIDE-REINFORCED
ALUMINUM ALLOY COMPOSITES

A Thesis
Presented for the
Master of Science
Degree
The University of Tennessee, Knoxville

Catherine Jessie Swindeman
August 1992

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DEDICATION

This thesis is dedicated to the Swindeman and Dwyer families

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ABSTRACT

This research involved the modification of conventional fusion welding techniques in an effort to successfully join discontinuous silicon carbide particulate-reinforced aluminum alloy composites. Based upon a preliminary evaluation of the weldability of several Al-matrix composites with respect to the gas tungsten arc welding (GTAW) and laser beam welding (LBW) processes, a mechanical assembly was designed, constructed, and modified for use during GTAW. Testing of the assembly involved GTA welding in the butt and spot modes with respect to various specimen geometries and welding parameters.

It was concluded that the modified fusion welding technique can be used to successfully join SiC/Al during GTA welding in the spot mode, although further work will be required in order to ensure repeatability. Additionally, it was shown that by controlling high temperature exposure during welding, a fusion zone can be produced in which the basic engineered properties of the composite are retained throughout the weld region.

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CHAPTER I

INTRODUCTION

Metal-matrix composite systems

Metal-matrix composites (MMCs) are metal-based, multi-component materials engineered to achieve a combination of properties that maximize the attributes of the individual constituents. Generally consisting of a dispersion of continuous or discontinuous fibers (f), whiskers (w), or particulates (p) in a metallic matrix, these materials offer the ability to match a wide variety of alloys and reinforcements. By modifying the type, orientation, and amount of reinforcement within the matrix, a composite can be tailored for a particular application. Where specific combinations of properties are desired, MMCs offer many advantages relative to monolithic materials. Improved mechanical and tribological properties can be achieved, in addition to physical characteristics which can be adjusted to meet design requirements.

Some of the advantages of MMCs as compared to their monolithic counterparts are [1]:

- higher stiffness/density ratio
- higher strength/density ratio
- improved wear resistance
- improved mechanical properties at elevated temperature
- adjustable physical properties, such as coefficient of thermal expansion and diffusivity

The potential for these improvements depends upon the particular filler/metal combination and the arrangement of the reinforcing constituent within the matrix. The addition of small, granular fillers (particulates) to a metal binder generally results in a material with higher specific stiffness than the matrix material, but with only a marginal improvement in strength. With whisker fillers, strength is slightly improved relative to particulates due to the higher aspect (length-to-diameter) ratio of whiskers which allows better "filler-to-filler" load transfer through the matrix [2]. Continuous fiber reinforcement optimizes the properties of the filler, achieving high specific strength and stiffness as a result of fiber continuity. Components designed to withstand forces applied in the direction of continuous fibers utilize the highest structural performance of the MMCs.

Interest in continuous fiber reinforced MMCs was stimulated in the early 1980's by the need for high temperature materials for aerospace and advanced engines, as well as materials with a high degree of thermal dimensional stability for applications such as space antennas [3]. Because of high fabrication costs, these types of MMCs were developed primarily for aerospace and military purposes. Currently, the highest production costs are still associated with continuous reinforcement [4], as shown in Figure 1. Due to the brittle nature of continuous ceramic fibers, net shape fabrication techniques such as diffusion bonding, powder metallurgy, and liquid metal infiltration are utilized in order to prevent fiber breakage [2,5-8].

A variety of advanced techniques (particularly powder metallurgy) are also available for producing discontinuous particulate or short whisker-reinforced MMCs. However, it is the ability to fabricate these composites by conventional foundry processes that is now

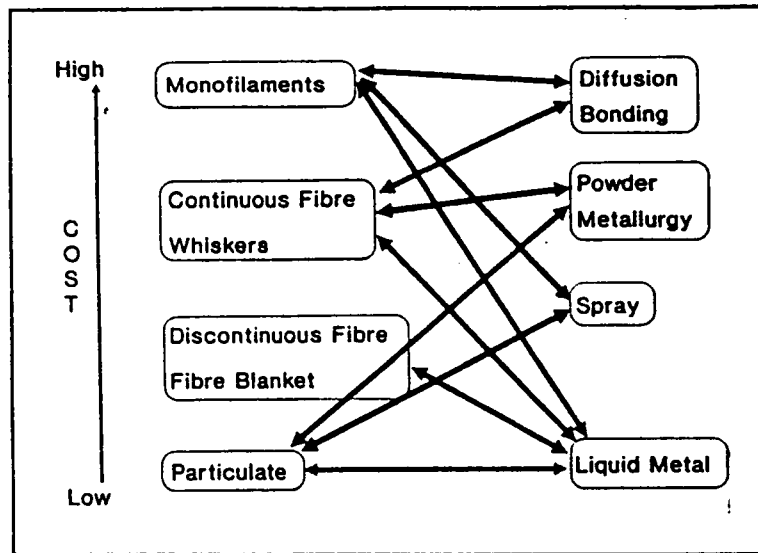


Figure 1. Relative costs associated with type of composite reinforcement and method of processing.

Source: Johnson, W.S. Metal matrix composites: Their time to shine?, *ASTM Standardization News*, October 1987, 36-39.

expanding their applications to industries where high production rates and low cost are required. The automotive industry, for example, has found numerous potential applications for particulate-reinforced metal composites [5]. Many of these components can be produced very economically by casting and standard metal working practices. Since particulates have a low aspect ratio, machinability and formability are facilitated relative to whiskers or fibers. Additionally, design capability with respect to particulate-reinforced MMCs is enhanced as a result of their almost isotropic properties [1,3].

Silicon carbide-reinforced aluminum alloy systems

Aluminum-based composites are currently the only MMCs available in mass production, due to their favorable cost and fabricability relative to other composites which incorporate light metals such as magnesium and titanium [1]. Commonly utilized matrix alloys include the 2XXX, 6XXX, and the 7XXX wrought aluminum alloy series, for which optimum strength levels are achieved by precipitation hardening [9], and casting alloys in the 3XX series. The selection of the reinforcing constituent is based on its cost and compatibility with the aluminum alloy matrix. In addition to the mechanical integrity and the morphology of the particular reinforcement, achieving a desirable combination of properties depends on such factors as the wetting characteristics of the filler, and its tendency to excessively react with the matrix at various temperatures.

In general, ceramics (carbides, oxides, and nitrides) are the preferred reinforcements for aluminum, due to their high specific strength and stiffness over a range of temperatures. Typical ceramic fillers for aluminum-based MMCs are silicon carbide (SiC), silicon nitride (Si₃N₄), and alumina (Al₂O₃). For continuous fiber-reinforced

aluminum alloys, SiC and Al₂O₃ are currently used, in addition to the earlier-developed aluminum-boron and aluminum-graphite systems. Alumina and alumina-silica are leading fiber materials for the discontinuous composites, while SiC is the primary whisker material [10].

Overall, SiC is the most common reinforcement for aluminum alloys [11]. The frequent use of SiC can be attributed to its high intrinsic strength and stiffness, high temperature stability, excellent oxidation resistance, and relatively high thermal conductivity and low coefficient of thermal expansion compared to many other ceramic materials [12]. SiC also offers the advantage of compatibility with Al alloys up to 500 °C [13], and its cost is moderate [2]. In addition to its capability as a continuous fiber and discontinuous whisker/platelet reinforcement, the versatile ceramic is seeing widespread use in the rapidly growing and relatively cost-effective area of particulate-based composites.

Mechanical properties of SiC/Al composites

Mechanical properties of continuously and discontinuously-reinforced SiC/Al systems illustrate the wide variety of improvements that can be achieved over unreinforced alloys. For example, unidirectional tensile strength of 1273 MPa was measured for a continuous SiC(f)-reinforced 6061 alloy [14], as compared to a tensile strength of 310 MPa for unreinforced 6061-T6 [15]. In tests on discontinuous SiC/Al, composites consisting of up to 40 volume % SiC whisker, nodule, or particulate-reinforced 6061, 2024, 2124, 7075, and 5083 demonstrated up to a 60 % increase in yield and tensile strength relative to the unreinforced “matrix” alloys [16]. The results of these tests (on the discontinuous composites) indicated that the yield strength, tensile strength, and the ductility were

controlled primarily by the matrix alloy, temper condition, and reinforcement content, where the highest properties were attained in SiC/6061. Additionally, the SiC/Al systems demonstrated increases in stiffness of 50-100 % over that of the unreinforced aluminum alloy matrices, where the modulus/density ratios were shown to be almost twice those of titanium and aluminum structural alloys. The elastic modulus of the discontinuous composites was found to be isotropic, independent of the type of reinforcement, and controlled only by the volume percentage of SiC reinforcement.

A comparison of mechanical properties of SiC/Al systems as evaluated by various researchers is shown in Table 1.

Applications of SiC-reinforced Al alloys

The properties achievable for continuous and discontinuous SiC-reinforced aluminum alloys have promoted their development for aerospace, structural, and automotive applications. In the aerospace industry, where performance generally takes precedence over cost, the high specific mechanical properties of more costly continuously reinforced SiC(f)/Al have resulted in their utilization as lightweight structural components. An example is the fabrication of SiC(f)/6061 for wing structural elements, where the "zee"-shaped wing stiffeners shown in Figure 2 are made by hot molding plasma-sprayed aluminum/fiber preforms into the desired shape [2].

For a Lockheed "next generation" fighter vertical tail, both continuous SiC fibers and discontinuous SiC whiskers in an aluminum alloy have been slated to constitute the skin and spars of the tail, respectively [18]. Demonstration aircraft parts also have been

TABLE 1
Mechanical properties of unreinforced and
SiC-reinforced aluminum alloys

Matrix alloy	Type (p or f), volume % SiC	Ultimate tensile strength (MPa)	Yield strength (MPa)	% elongation	Elastic modulus (GPa)	Reference number
A356-T61	0	255	200	4.0	75.2	20
A356-T61	p,10	276	262	0.7	77.2	20
A356-T61	p,15	303	296	0.4	92.4	20
A356-T61	p,20	317	296	0.5	95.8	20
6061-T6	0	310	275	12-17	69.0	15
6061-T6	p,30	440	-	1.5	117	14
X2080-T6	p,15	527	436	4.0	102	17
X2080-T6	p,20	548	464	4.0	>112	17
7075-T6	p,30	613	-	0.9	126	14
6061	f,47	1273	-	0.9	204	14

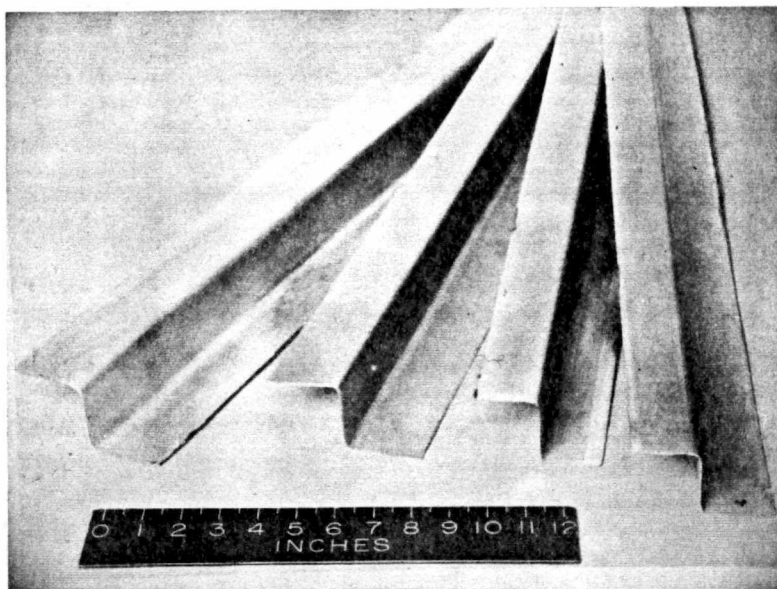
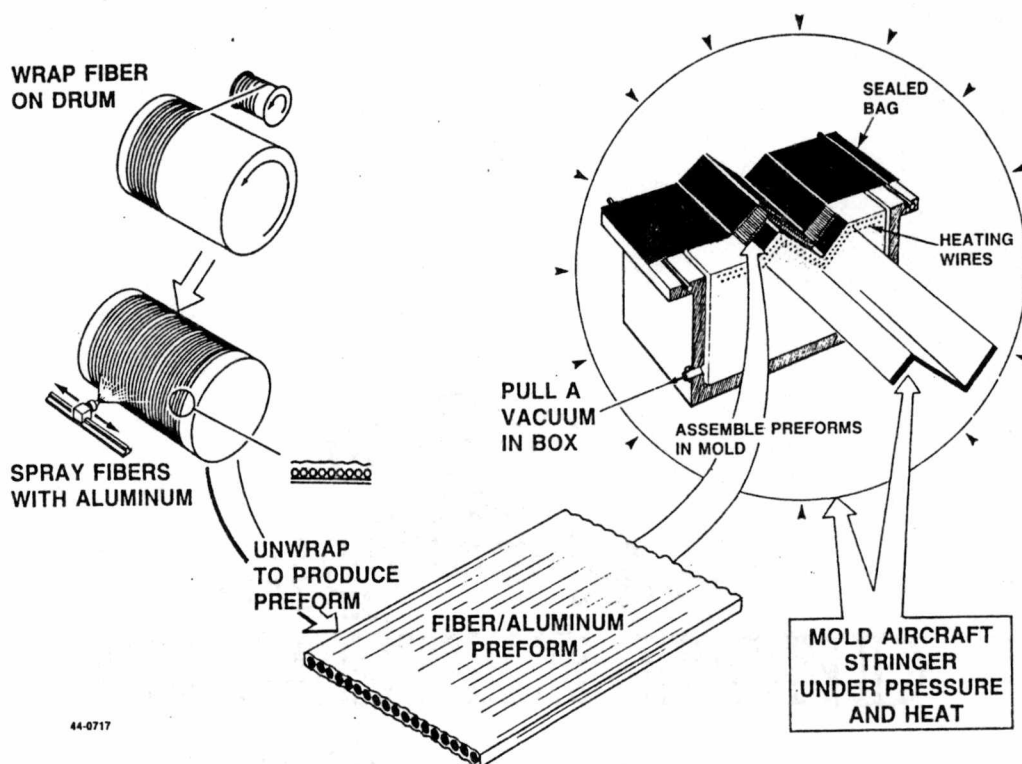


Figure 2. SiC(f)/Al "Zee" panel stiffeners fabricated by hot molding.

Source: Textron Specialty Materials, *Continuous Silicon Carbide Metal Matrix Composites*, 2 Industrial Avenue, Lowell, MA 01851.

recently fabricated using SiC particulate reinforced alloys 2124 and 8090, where composite billets were prepared by powder metallurgy, and then processed by "conventional methods" into sheet, plate, extrusions, or forgings to make such products as missile control fins [19]. Currently, composites such as SiC(p)-reinforced A356 and A357 are being fabricated by investment casting for aircraft camera gimbals and are under testing for brackets and missile parts (Figure 3) [20].

One area in which the "tailorability" of SiC/Al has been fully utilized is electronic packaging for avionics. Electronic components for aircraft and spacecraft, such as ceramic circuit substrates, require structural support, environmental protection, and heat removal; so, suitable packaging materials must offer the combination of low weight, good corrosion resistance, good thermal conductivity, and dimensional stability. Additionally, these delicate components must have thermal expansion characteristics that are similar to their housings, in order to prevent failure during thermal cycling [11].

SiC(p)/Al and SiC(w)/Al offer all the necessary attributes for packaging applications. By controlling the amount of reinforcement, the coefficient of thermal expansion (CTE) of the composite can be tailored to match the low CTE of electronic circuits. Thus, SiC(p)/Al has been developed as a replacement for Kovar in microwave circuit packages by virtue of its lower density, higher thermal conductivity, and similar CTE, as well as other superior properties. In one instance, substitution of a 40 volume % SiC(p)/6061 powder metallurgy-consolidated composite for Kovar in a microwave package achieved a 63% weight savings and a six-fold increase in thermal conductivity [21]. A comparison of the SiC/Al composite's properties to those of Kovar and other packaging materials is shown in Table 2 [22].

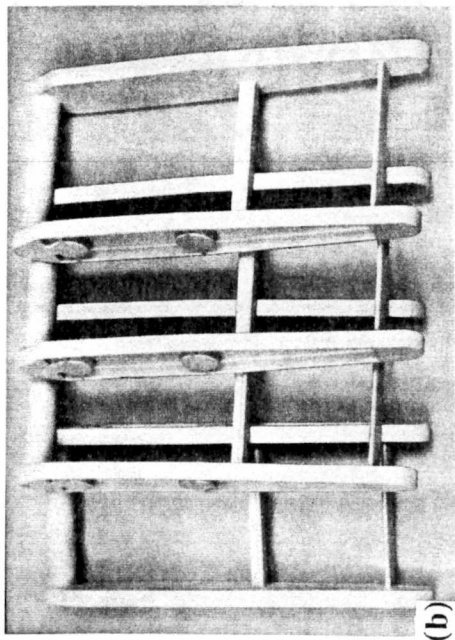
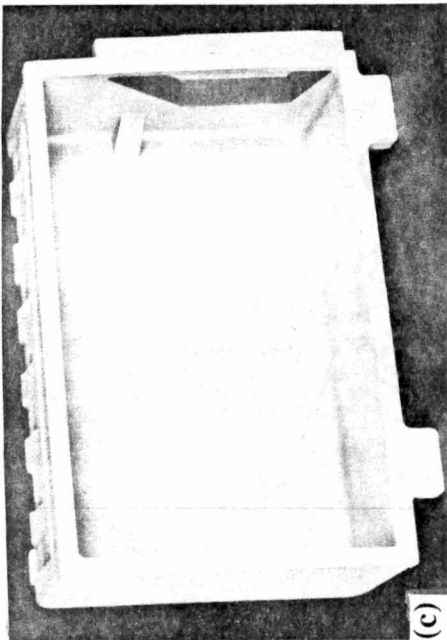
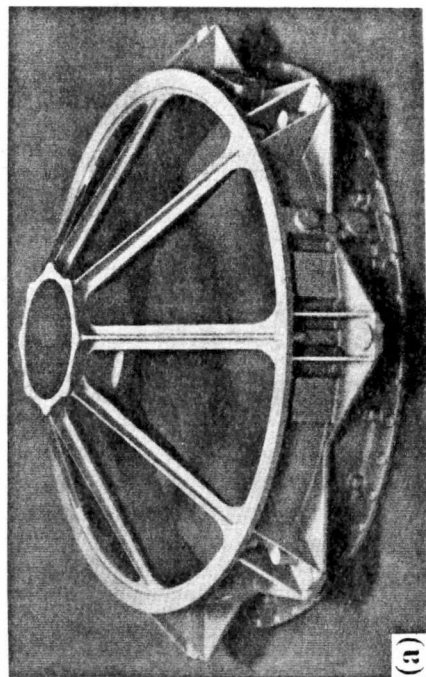


Figure 3. Cast 10 to 20 volume % SiC(p)/A356 aircraft-camera gimbel (a), brackets (b), and pressure vessel component (c) [20].

Source: Kennedy, David O., SiC particles beef up investment-cast aluminum, *Advanced Materials & Processes*, June 1991, 42-46.

TABLE 2
Comparison of thermal properties and densities of materials

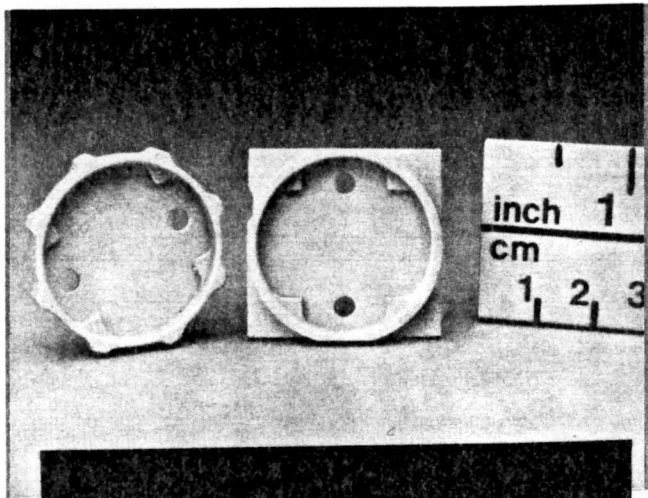
	Density (lb/in³)	Thermal conductivity (W/m/K)	Specific thermal conductivity (K/ρ)	C.T.E. (ppm/F)
Aluminum	0.098	221	2255	13.0
Copper	0.320	391	1220	9.7
Molybdenum	0.366	146	400	2.9
Aluminum Nitride	0.117	180	1538	1.9
Alumina	0.129	22-36	170-279	3.6-4.2
Copper/Invar /Copper	0.293	131	447	3.0
Kovar	0.293	17	58	3.0
50 - 55 vol % SiC(p)/Al MMC	0.108	200-220	1850-2040	4.4-4.8

Source: DWA Composites Specialties, Inc., 21119 Superior Street, Chatsworth, CA, 91311-4393.

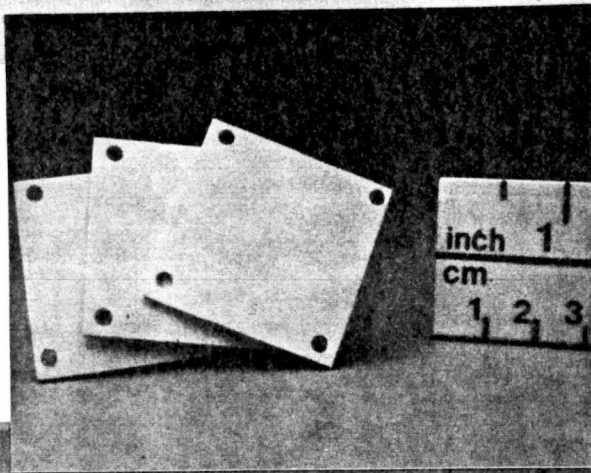
As a result of these performance advantages, SiC/Al alloys were also approved to replace beryllium for instrument covers in missile guidance systems [23]. The composites achieved a CTE closely matching that of Be, while having a higher specific stiffness and eliminating the toxicity hazard associated with processing beryllium. SiC/Al components have also been employed in electronic applications such as heat sinks and heat pipes [11], where SiC(p)/A357 heat sinks and other packaging components are shown in Figure 4 [24].

The attributes of SiC particulate-reinforced Al alloys have stimulated significant efforts related to the utilization of these materials by the automotive industry. As mentioned earlier, this interest was fueled by the development of affordable fabrication methods for high-volume production automotive parts [25]. Ingots of SiC(p)/Al alloys are remelted and cast by sand casting, permanent mold casting, investment casting, expendable-pattern casting, squeeze casting, and die casting to make pistons, brake rotors, cylinder sleeves, and drive shafts [5]. The use of SiC(p)/Al in diesel-engine pistons, for example, reduces thermal expansion and improves elevated temperature strength relative to unreinforced aluminum; critical areas of the squeeze-cast piston shown in Figure 5 were strengthened by selective reinforcement of crown and ring lands [26]. According to one researcher, it is expected that such reinforced aluminum pistons will equip virtually every new diesel truck within the early 1990's [27].

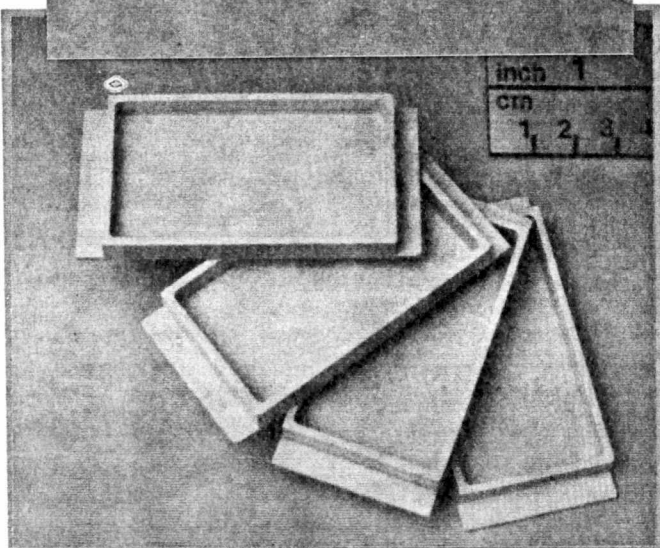
With the continued success of low cost ingot-metallurgy fabrication processes, SiC(p)/Al composites may expand from aerospace and automotive applications into many



Solid state sensor housings
A357 + 20 % SiC(p)



Heat sinks
A357 + 60 % SiC(p)



Electronic packages
A357 + 20 % SiC(p)

Figure 4. Cast SiC particulate-reinforced Al alloy composites.

Source: Kennerknecht, S., MMC studies via the investment casting process, *Conference Proceedings: Fabrication of Particulates Reinforced Metal Composites*, Materials Park, Ohio: ASM International, 1990, 87-100.

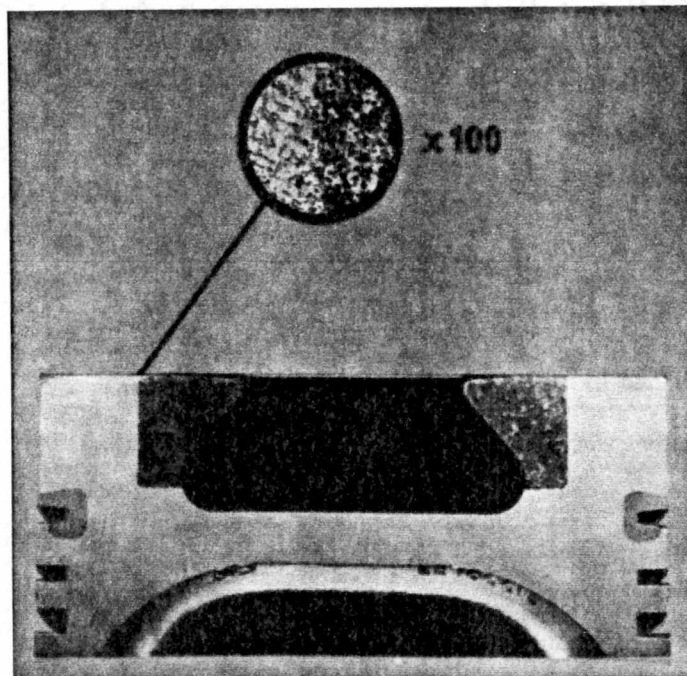


Figure 5. Cast automotive piston incorporating SiC/Al.

Source: Lampman, Steven R., Tuning up the metals in auto engines, *Advanced Materials & Processes*, May 1991, 17-22.

other areas. Potential applications that have been mentioned vary from pressure vessels to medical prostheses and golf-club heads [20].

Joining of SiC/Al

The continued utilization and/or possible high-volume production of SiC-reinforced aluminum alloys will depend on many factors, one of which is the development of a database pertaining to mechanical properties, corrosion sensitivity, metallurgical processing, and welding/joining techniques. An area in which a relatively small amount of work has been carried out is welding of SiC/Al alloys. Unfortunately, this is probably related to the fact that attempts to join SiC/Al by conventional welding techniques have been largely unsuccessful, which threatens to limit the potential applications of these composites.

The desire to utilize fusion welding processes is based on convenience, economy, and the potential to achieve the full benefits of the engineered properties of the composite. However, the major problem associated with these processes in relation to SiC-reinforced composites - as with most other joining techniques that have been attempted - is the inability to retain the basic characteristics of the composite structure in the joint. Fusion welding of SiC/Al has resulted in degradation of composite properties corresponding to the loss of the silicon carbide reinforcement and the formation of an undesirable reaction product in the fusion zone, which is due to the reaction of the reinforcing SiC with the aluminum alloy matrix.

Efforts by various researchers to fully inhibit the reaction between the matrix and reinforcement that occurs during welding of SiC/Al by "conventional" fusion welding

processes have met with little success. These attempts have indicated, however, that the extent of reaction can be significantly reduced by minimizing heat input during welding, and that SiC dissolution and the formation of the undesirable reaction product can be limited to a portion of the fusion zone in which peak temperatures are encountered. Based on this evidence, it may be possible to produce weldments having desirable properties if this portion of the fusion zone can be eliminated by some means.

While it has been reported that techniques other than autogenous fusion welding have achieved successful results in joining SiC-reinforced Al alloys, no alternative method has been described in the available literature as possessing the capability to join average-size components while maintaining the distribution of SiC necessary for retention of the engineered properties throughout the composite. In spite of the problems associated with autogenous fusion welding, it remains perhaps the best potential means of producing a joint in which the composite properties are retained. Fusion welding will additionally be the most economical method of joining SiC/Al, if it can be successfully developed with respect to these MMCs.

Objective of study

Based on the potential advantages of fusion welding, and the results of previous research indicating the ability to minimize the matrix-reinforcement reaction, it was of interest to further investigate the possibility of overcoming the metallurgical problems associated with fusion welding of SiC/Al composites. Initially, the purpose of this study was to verify some of the results of the relatively limited number of previous studies involving fusion welding of SiC-reinforced Al alloys. After this verification was

accomplished, the objective was to determine whether a modification to a conventional fusion welding process could be implemented in order to produce improved results. An assessment of previous studies indicated that an approach that had not yet been attempted on a significant scale was the elimination of a portion of the fusion zone by mechanical means. Thus, the goal of this study was to devise a method of applying pressure during fusion welding in order to eliminate the portion of the weld pool associated with filler/matrix interaction, leaving the region in which SiC particles remain intact to solidify. This was to be accomplished by the construction of a mechanical device that could potentially be used with various fusion welding processes.

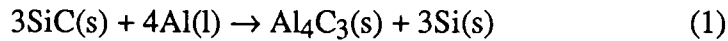
CHAPTER II

LITERATURE REVIEW

Filler-matrix interaction during fabrication

Two of the primary concerns related to the development of SiC-reinforced aluminum alloys are the dissolution of the SiC reinforcement and its reaction with the matrix to form an undesirable compound - occurrences that are due to the thermodynamic instability of SiC in molten aluminum. This instability has been observed by numerous researchers studying casting or other methods of fabricating SiC/Al, who noted the attack of reinforcing SiC by molten aluminum to form a brittle aluminum carbide compound [28-35].

Ternary phase diagrams for the Al-C-Si system have been developed by Warren and Andersson [36], Viala, et al [37], and Oden and McCune [38], as shown in Figures 6 through 8. Based on analysis of the diagram established by Viala, et al, Laurent, et al [32] concluded that since no two-phase field exists with silicon carbide and pure liquid aluminum in equilibrium, SiC and Al should react to form aluminum carbide, Al_4C_3 , and an Al-Si alloy above the liquidus of aluminum, where the reaction proceeds as:



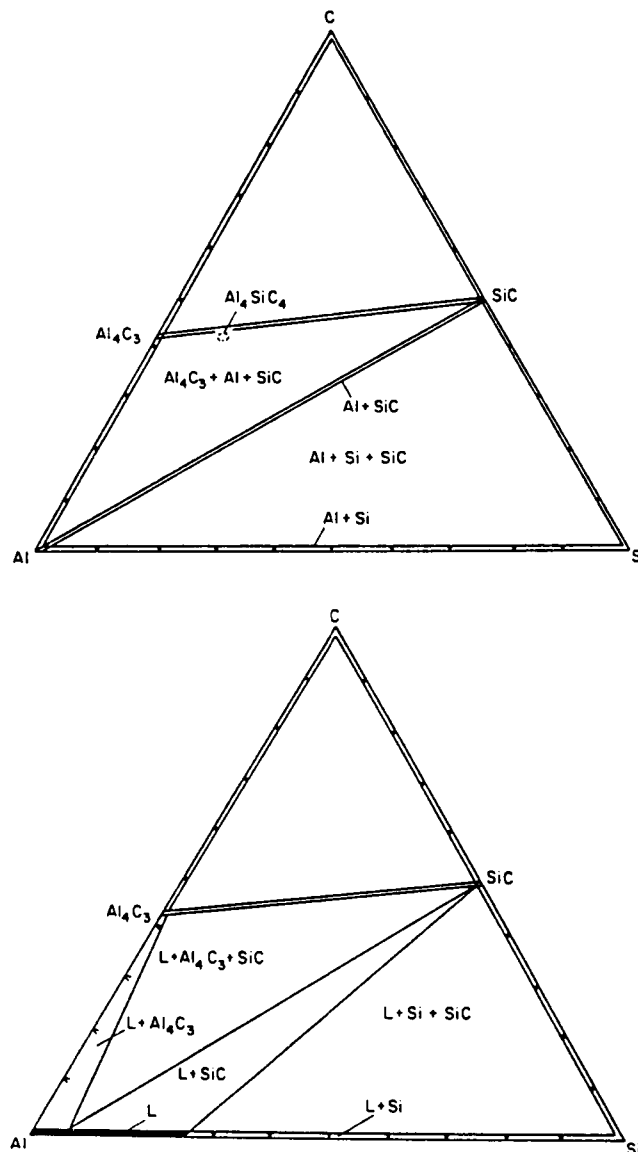
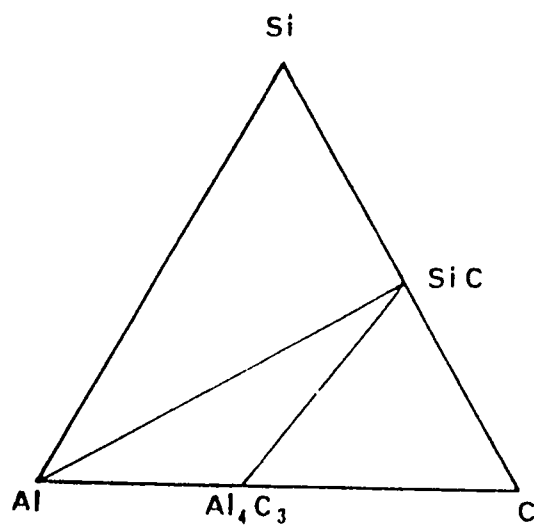


Figure 6. The Al-Si-C system at 500 °C (top) and 800 °C (bottom).

Source: Warren, R., and Andersson, C.H., Silicon carbide fibres and their potential for use in composite materials, Part 2, *Composites* 15 (2), April 1984, 101-111.

a - 830 K



b - 845 K

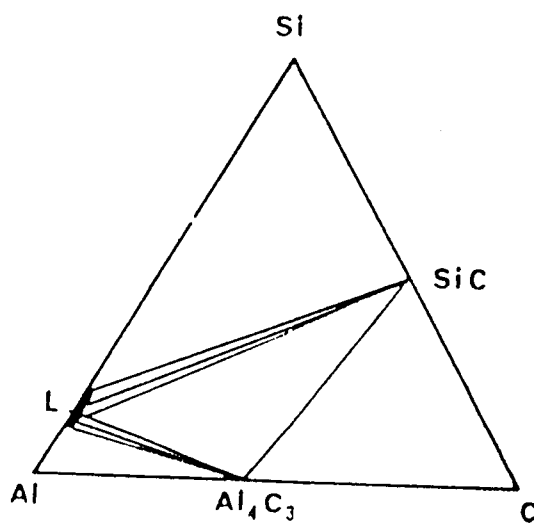


Figure 7. The Al-C-Si system at 830 K (a) and 845 K (b) (2000 MPa).

Source: Viala, J., Fortier, P., Bonnetot, B., Bouix, J., High pressure interaction between solid Al and SiC, *Materials Research Bulletin* 21, 1986, 387-394.

While the occurrence of this reaction is obviously undesirable in relation to the dissolution of the reinforcing SiC, there are other negative aspects associated with the presence of Al_4C_3 :

(a) the formation of the carbide at the reinforcement/matrix interface may disturb interfacial properties by interfering with the effective transfer of stress between the matrix and reinforcement, thereby degrading mechanical properties [34].

(b) corrosion susceptibility is increased due to the chemical properties of aluminum carbide. In addition to decomposing in aqueous environments and with heating in the presence of oxygen and oxygenated compounds such as metal oxides, the carbide is dissolved by solutions of acids and alkalis with evolution of methane. At elevated temperatures (above about 650 °C), it is attacked by chlorine and bromine [39]. Corrosion in humid environments is due to hydration of aluminum carbide and the solution of Al with the formation of hydroxides in a galvanic couple of Al-C formed when moisture penetrates to the interface [40].

(c) Al_4C_3 is a very hard compound, and thus in significant quantity may impart high hardness and brittle behavior to the material containing it. The hardness of the carbide has been compared to that of topaz, which is designated as 8 on the Mohs scale of 1 to 10, where diamond is 10 [41].

(d) the reaction which produces the aluminum carbide also increases the silicon content of the alloy, thus markedly altering the matrix composition if extensive reaction occurs [42].

Additionally, Al_4C_3 is said to be stable up to approximately 2100 °C [39].

An expression for the Gibbs free energy (ΔG) of Equation (1) was determined by Isaikin, et al [43] and rewritten by Iseki, et al [44]. It is shown as Equation (2):

$$\begin{aligned} \Delta G(\text{J/mol}) = & 113900 - 12.06T \ln T + 8.92 \times 10^{-3} T^2 \\ & + 7.53 \times 10^{-4} T^{-1} + 21.5T + 3RT \ln a_{[\text{Si}]}, \end{aligned} \quad (2)$$

where $a_{[\text{Si}]}$ is the activity of silicon in liquid aluminum. Using the variation of $a_{[\text{Si}]}$ with silicon content for the Al-Si system, Isaikin, et al, calculated equilibrium concentrations of Si in Al in contact with SiC. The minimum Si content considered was 7.5 at %, where the calculations predicted conditions of equilibrium at a temperature of 1000 K. Thus, for this matrix composition, the formation of the reaction products in Equation (1) should not occur below about 730 °C (1350 °F) because ΔG is positive. But above this temperature, the free energy change becomes increasingly negative, and the reaction proceeds. Iseki, et al calculated free energy changes based on Equation (2), noting that as the amount of silicon in liquid aluminum increases and $a_{[\text{Si}]}$ becomes greater, the reaction tends to saturate. By increasing the initial silicon content in aluminum, the reaction reached the saturation level sooner; that is, a prior addition of silicon to the matrix alloy inhibited the formation of Al_4C_3 .

The inhibiting effect of Si on the SiC/Al reaction can also be predicted on the basis of the ternary phase diagram developed by Warren and Andersson shown in Figure 6 [36]. In contact with SiC at a typical casting temperature of 500 °C, pure aluminum would be

expected to react to some extent with the ceramic - since the section Al-SiC is adjacent to the three phase field corresponding to the presence of Al_4C_3 , Al, and SiC. Increasing the silicon content of the alloy, however, disallows the formation of aluminum carbide upon heating to 500 °C, since the nominal alloy composition lies in the Al-Si-SiC region.

Based on the ternary phase diagram developed by Viala et al, Laurent, et al [32], noted that the silicon content in the matrix alloy required to avoid reaction of SiC and liquid aluminum must correspond to the three-phase equilibrium of $(\text{Al-Si})_{\text{liquid}}$, Al_4C_3 , and SiC. This silicon content of the Al-Si alloy in equilibrium with silicon carbide and aluminum carbide increases with temperature, as determined by Viala, et al (Figure 9, [37]).

Methods of inhibiting carbide formation

For thermodynamic stability at typical casting temperatures, levels of silicon required to discourage the formation of aluminum carbide have been calculated to be about 10 to 12 weight percent; however, these values may be reduced depending upon the holding time at a particular temperature. The shorter the time the material is held at this temperature, the lower the necessary amount of silicon. For the hypoeutectic Al-Si-Mg casting alloy A356, which contains 6 to 7 wt% Si and 15 vol % SiC, "barely detectable" levels of Al_4C_3 were produced for holding at 750 °C for 4 hours, illustrating that the rate of aluminum carbide formation is quite slow near the melting point of this alloy. The slow rate of Al_4C_3 formation was based on X-ray diffraction results for SiC(p)/A356, which showed a gradual increase in silicon intensity with time (Figure 10, [36]).

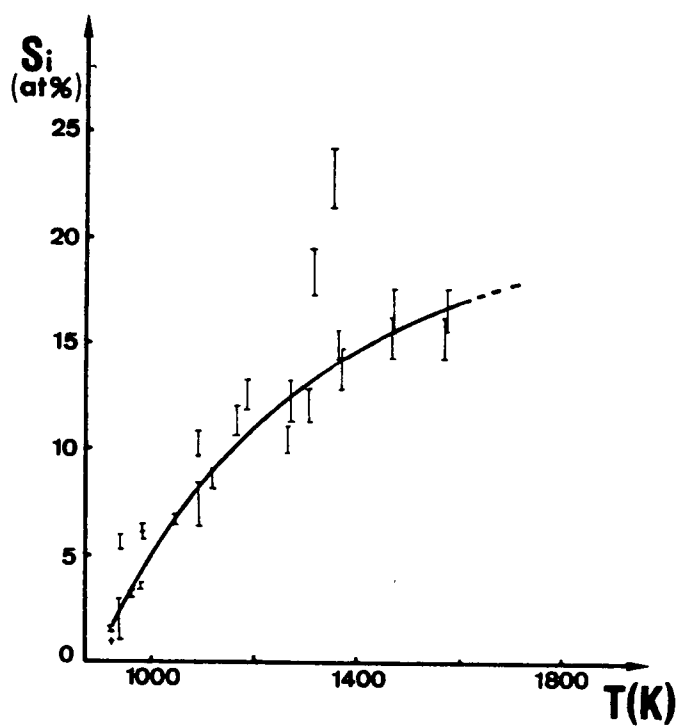


Figure 9. Variation in silicon content of the Al-Si alloy in equilibrium with SiC and Al₄C₃ as a function of temperature.

Source: Viala, J., Fortier, P., Bonnetot, B., Bouix, J., High pressure interaction between solid Al and SiC, *Materials Research Bulletin* 21, 1986, 387-394.

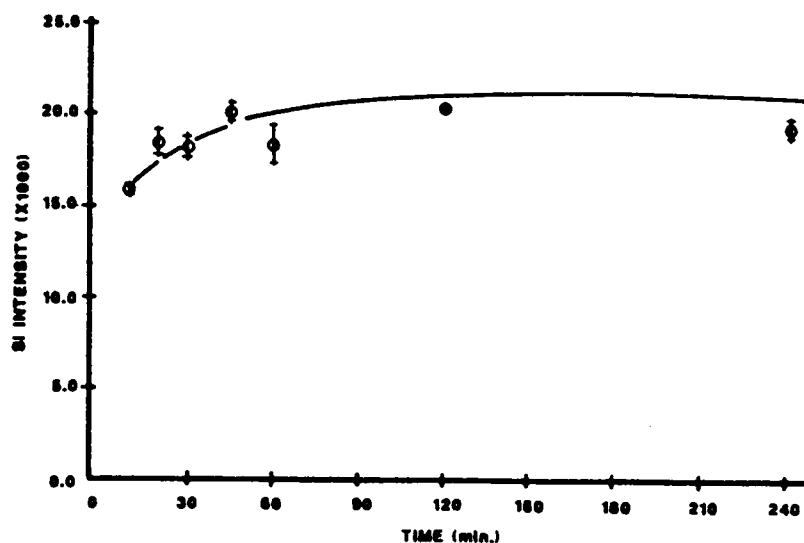


Figure 10. X-ray diffraction results for 15 volume % SiC(p)/A356 at 800 °C.

Source: Warren, R., and Andersson, C.H., Silicon carbide fibres and their potential for use in composite materials, Part 2, *Composites* 15 (2), April 1984, 101-111.

In the most important commercially cast aluminum alloys, silicon - which enhances fluidity and thus castability - is already present [45] in amounts which may inhibit significant reaction between SiC and the aluminum-based matrix as just described. Highly castable Al-Si and Al-Cu-Si alloys typically contain from 5 to 13 wt % Si, and up to 22 wt % Si, respectively [46]. Featuring silicon contents of about 7 wt % [45], the alloys A356 and A357 are becoming the most popular matrix materials for cast SiC-reinforced composites as a result of their castability and competitive cost. For example, Kennedy [20] recently noted the start-up of a Duralcan plant which is expected to produce SiC(p)-reinforced A356 and A357 at the rate of 12,500 tons/year.

In a study on squeeze cast composites, Diwanji and Hall [33] compared the interfacial characteristics of 35 vol % carbon fiber-reinforced A356 to various carbon fiber-reinforced Al-Mg and Al-Mg-Zn composites. Extensive T.E.M. analysis of interfacial areas showed the presence of carbides in all composites. Through convergent beam diffraction patterns and micrographs taken under lattice imaging conditions, the carbides were identified as Al_4C_3 , where the silicon-containing A356-based composites exhibited the minimum amount of these carbides. Subsequent EDS line scans of the matrices confirmed concentration gradients of silicon in the vicinity of the fibers in A356, and micrographic evidence showed significant portions of second phase silicon at the interfaces. These observations led to the conclusion that, aside from the carbide formation, there was no other reaction occurring between the fibers and various matrices and that the presence of silicon in the matrix restricts the formation of Al_4C_3 .

The drawback of Al-Si cast aluminum composites, as with unreinforced aluminum alloys, is that they tend to exhibit inferior strength relative to those alloys with lower levels

of silicon [47]. For example, an ultimate tensile strength of about 850 MPa was reported for the 35 vol % carbon fiber-reinforced A356 composite in the study just described, while a UTS of 1110 MPa was measured for the Al-3Mg-2Zn-based composite [33]. The reduced strength associated with higher Si content is also shown in Table 1, where composites having A356-T61 matrices exhibited a lower UTS than the low-Si composites incorporating 6061, X2080, and 7075 matrices in the “T6” condition. Thus for applications which require higher strength, low silicon aluminum alloys have been employed as matrix materials.

Matrices of heat treatable wrought alloys such as 2014, 2024, and 6061 have been frequently evaluated in combination with silicon carbide reinforcement. But since these alloys generally contain much less silicon than the 6 to 7 wt% calculated to prevent significant filler/matrix interaction (2014, 2024, and 6061 have less than 1 wt % Si), more extensive carbide formation can be expected to occur at elevated temperature and shorter holding times. In addition to SiC dissolution and carbide formation, the modification of the chemical composition due to excess silicon is a disadvantage in that matrix properties of high strength wrought alloys can be significantly degraded [34].

In composites having low silicon matrices, property degradation must be minimized by methods such as controlling reaction kinetics by limiting high temperature exposure or using protective coatings on SiC reinforcements. For example, Legoux, et al [28], observed no interfacial relation between SiC and low silicon-alloy matrices such as 6061, when composites were fabricated by a compocasting process. Inhibition of the reaction was probably due to the relatively short times at the low temperatures associated with compocasting, a procedure which involves adding reinforcing SiC particles to an agitated,

semi-solid alloy slightly below its liquidus temperature, followed by remelting and solidifying under pressure [13].

The effect of various coatings on SiC reinforcements has also been studied in relation to reducing the interfacial reaction rate. Teng and Boyd [29] applied thin, continuous coatings of Al_2O_3 and MgO on SiC particulates for reinforcement of both A356 and low silicon-alloy matrices and evaluated interfacial reaction relative to uncoated SiC/Al matrix combinations. Following fabrication by squeeze casting, A356-based composites were held at 800 °C for up to 8 hours. Metallographic examination showed large semi-continuous "clumps" of a Si-rich phase in MMCs containing uncoated particulates, while there was no evidence of this phase in the composites containing coated SiC for heat treating up to 5 hours; this suggests that the Si-rich regions were formed as a result of the dissolution of SiC particulates due to reaction with aluminum, and that the coatings inhibited this reaction. Low silicon-matrix composites were fabricated by liquid metal infiltration and then held at about 720 °C so that the time dependence of the interfacial reaction could be monitored. As with the cast composites, the extent of reaction was reduced where coated particulates were used.

The presence of a layer of silica on silicon carbide reinforcements has also been shown to limit the formation of Al_4C_3 and excess silicon. Legoux, et al [28], remelted low silicon-matrix composites at 800 °C for 1/2 hour and observed significantly less SiC dissolution with oxidized particulates. Better protection of SiC was attributed to the reaction of silica with the aluminum matrix to form spinel (MgAl_2O_4), although a volume contraction accompanying the reaction was thought to have left this layer of spinel "porous" enough to allow some liquid Al to migrate towards SiC and react with it.

Fusion welding

The reaction of SiC reinforcements with an aluminum-based matrix can be significantly reduced by the use of a high silicon matrix content, limited elevated temperature exposure, and the implementation of protective filler coatings. In many instances, it is not possible to take advantage of all of these factors, as fiber coatings mean higher fabrication costs, high silicon contents are not always appropriate, and high temperature exposure may be very difficult to limit. Desirable properties for SiC-reinforced low-Si aluminum alloy composites have been achieved nonetheless by exercising careful process control. However, retention of these properties (even for high Si matrices) has proven very difficult to accomplish during a secondary fabrication process such as welding, as the temperatures experienced in welding readily promote filler/matrix reaction. As shown in Figure 11[44], the free energy change associated with SiC-Al reaction for a constant activity of silicon becomes increasingly negative with higher temperature; that is, the effect of $a_{[Si]}$ decreases with increasing temperature. Thus, attempts to join SiC/Al alloy composites using several fusion welding processes have resulted in the decomposition of SiC and the formation of Al_4C_3 .

Gas tungsten arc welding

The most common "conventional" welding process utilized in attempts to join SiC/Al composites has been gas tungsten arc welding (GTAW) - a process that is extremely well suited for welding aluminum alloys. In GTAW, coalescence is produced by the heat of an arc generated between a nonconsumable tungsten electrode and the weld pool, and it

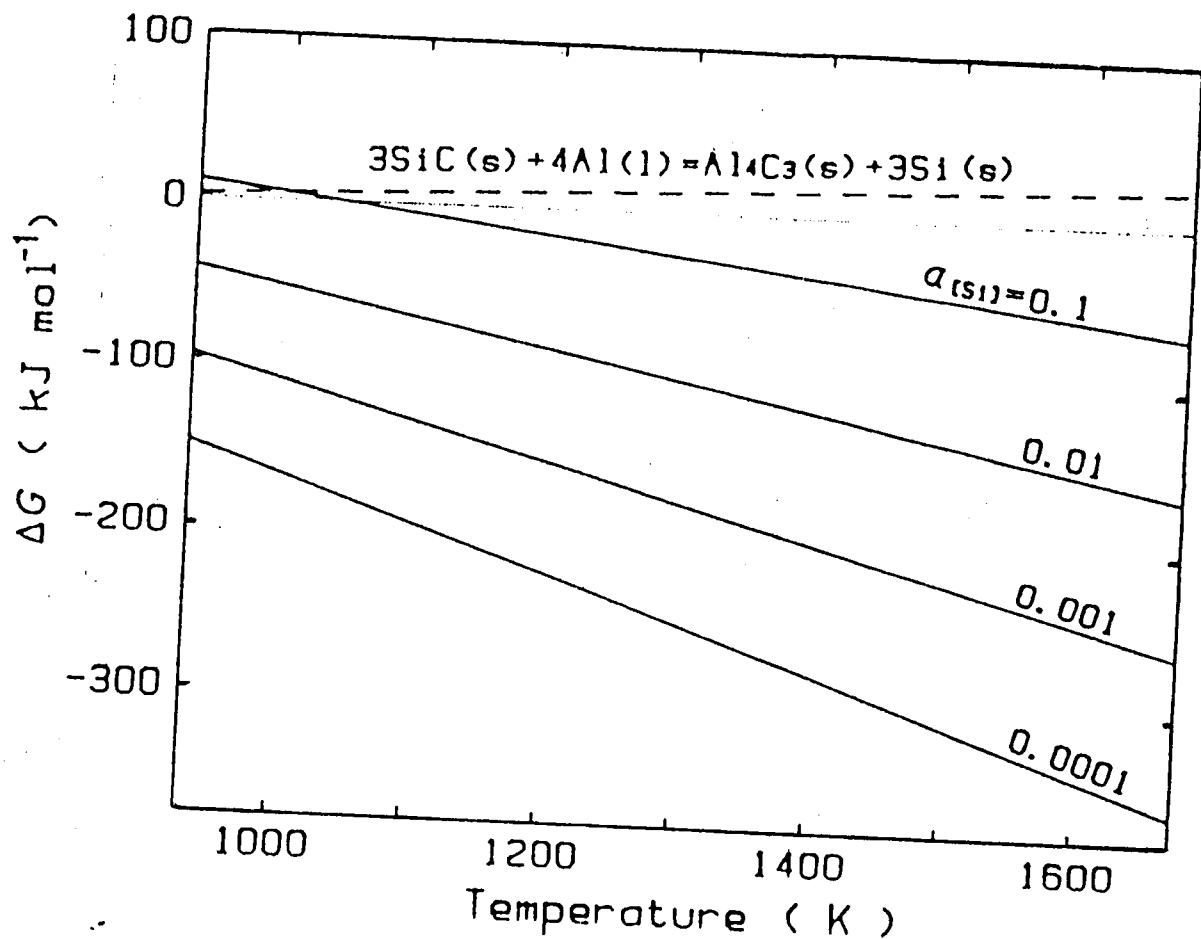


Figure 11. Free energy change of the SiC-Al reaction.

Source: Iseki, T., Kameda, T., and Maruyama, T., Interfacial reactions between SiC and aluminium during joining, *Journal of Materials Science* 19, 1984, 1692-1698.

is carried out with the use of shielding gas and with or without the addition of filler metal. Details regarding GTAW are described as follows [48]:

As shown in Figure 12 , the tungsten (or tungsten alloy) electrode is held in a torch, through which shielding gas is fed in order to protect the electrode, the molten weld pool, and solidifying weld metal from atmospheric contamination. An electric arc is produced by current passing through the conductive, ionized shielding gas, and the heat of this arc melts the base material. The GTAW process offers several advantages:

- (a) generally defect-free, superior quality welds are produced.
- (b) welding variables can be precisely controlled.
- (c) the process is spatter free, as opposed to other welding processes.
- (d) autogenous welds can be produced at high speeds, using relatively inexpensive power supplies.
- (e) filler metal may or may not be used. If it is used, it can be added independently of the heat source.

With respect to Al alloys, a primary advantage of GTAW is the surface cleaning capability it offers; one of the problems associated with welding of aluminum alloys is the formation of refractory surface oxides that inhibit wetting of

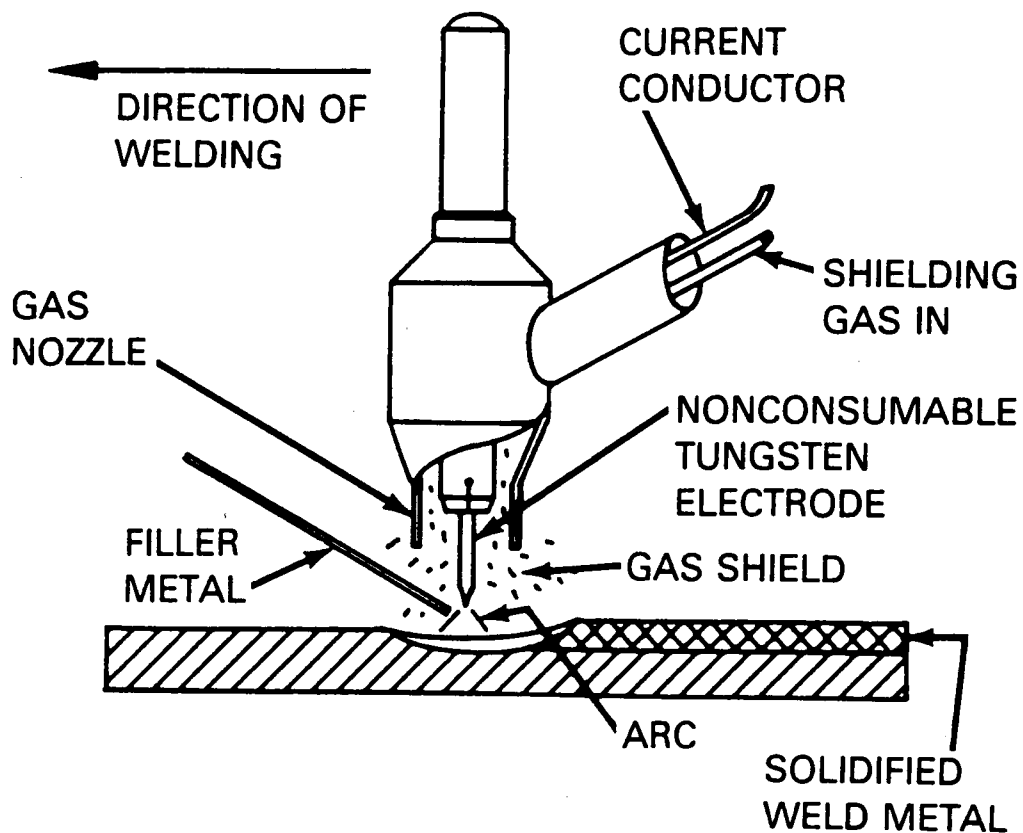


Figure 12. Gas tungsten arc welding operation.

Source: O'Brien, R.L., ed., *Welding Handbook*, Vol 2, 8th ed., Miami, Florida: American Welding Society, 1991, 74-107.

the weldment by the weld metal. In GTAW, this problem can be combatted by adjusting the welding parameters such that oxides can be removed by a cathodic cleaning action created at the workpiece surface. This cleaning action can be generated by using DCEP (direct current, electrode positive) during welding. As shown in Figure 13, the tungsten electrode is connected to the positive terminal of the power supply with DCEP, which results in the flow of electrons from the work to the electrode and the generation of approximately 70 % of the arc heat at the electrode and 30 % at the workpiece. Since the positive electrode is being heated by the bombardment of electrons and by its own resistance to their passage, the use of a larger diameter electrode for a given welding current (relative to that required if the electrode were negative) is necessary with DCEP in order to reduce this heating. DCEP is thus limited to welding sections having thicknesses that do not exceed about 1/8 inch.

Connecting the electrode to the negative terminal of the power supply for DCEN (direct current, electrode negative) results in 70 % of the arc heat being generated at the anode (work) and the remaining 30 % at the electrode. As a much greater amount of heat is concentrated at the work in this case, superior penetration is achieved and welding can be conducted at higher travel speeds. However, DCEN does not provide the cleaning action of DCEP. Therefore, it is frequently desirable to carry out welding using an alternating current with respect to aluminum alloys, which combines the cleaning capability of DCEP and the penetration characteristics of DCEN.

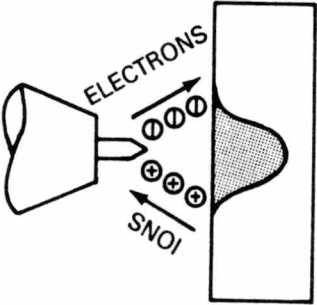
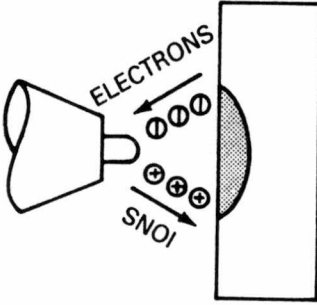
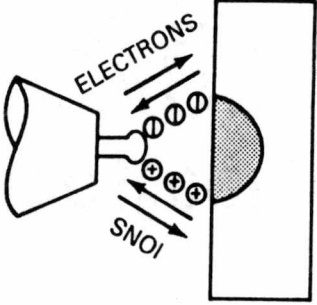
CURRENT TYPE	DCEN	DCEP	AC (BALANCED)
ELECTRODE POLARITY	NEGATIVE	POSITIVE	
  			
PENETRATION CHARACTERISTICS			
OXIDE CLEANING ACTION	NO	YES	YES-ONCE EVERY HALF CYCLE
HEAT BALANCE IN THE ARC (APPROX.)	70% AT WORK END 30% AT ELECTRODE END	30% AT WORK END 70% AT ELECTRODE END	50% AT WORK END 50% AT ELECTRODE END
PENETRATION	DEEP; NARROW	SHALLOW; WIDE	MEDIUM
ELECTRODE CAPACITY	EXCELLENT e.g., 1/8 in. (3.2 mm) 400 A	POOR e.g., 1/4 in. (6.4 mm) 120 A	GOOD e.g., 1/8 in. (3.2 mm) 225 A

Figure 13. Characteristics of current types for GTAW.

Source: O'Brien, R.L., ed., *Welding Handbook*, Vol 2, 8th ed., Miami, Florida: American Welding Society, 1991, 74-107.

Depending on the desired welding characteristics, an appropriate shielding gas is selected for GTAW of aluminum alloys. The most common types of inert gas used for GTAW shielding are argon and helium, or mixtures of these two gases. With argon, improved arc starting, better cleaning action, and reduced penetration are obtained. More expensive helium is better suited to welding thicker sections and materials with higher thermal conductivity, as it transfers a greater amount of heat to the work than argon. The flow rate of helium, however, must be two to three times higher than that of argon during welding in order to attain the same shielding effectiveness. Therefore, except where helium's higher heat penetration is required, argon is generally preferred for most applications.

Weldability studies of SiC/Al in relation to GTAW

SiC-reinforced aluminum alloys have been evaluated by a number of investigators in relation to GTAW. In an investigation carried out by Devletian [49], GTAW was utilized in order to weld 40 % SiC(p)-reinforced 6061 over a wide range of heat inputs. Metallographic inspection showed that at every level of heat input, the autogenous fusion welds were characterized by poor fluidity and gross porosity, in addition to the formation of plate-like precipitates and large blocky precipitates. While poor fluidity was attributed to the large volume fraction of solid SiC particulates in the pool of molten aluminum, the increasing volume of porosity that appeared as heat input increased was associated with the inability of gaseous bubbles to escape from the weld pool due to blockage by the SiC particulates.

Through electron microprobe microanalysis, the plate-like precipitates were identified as Al_4C_3 , and the blocky precipitates were identified as Si. The amount of aluminum carbide was observed to increase with heat input along with increasing alteration of the size and distribution of SiC particulates.

Ahearn, et al [50] also noted high porosity produced by GTA welding of SiC/6061, and attributed it to hydrogen absorbed by the parent 6061. In order to minimize pore density by promoting the diffusion of hydrogen to the sample surface, vacuum degassing was carried out by heating 15-20 vol % SiC(p)/6061 at various rates. Samples that were subjected to rapid thermal cycling from ambient temperature to temperatures exceeding 750 °C (to simulate welding) were characterized by significantly higher pore densities than samples that were slowly heated to the same temperatures. However, metallographic and X-ray structural analysis of the degassed samples - including those in which pores were eliminated - revealed that thermal cycling had resulted in the decomposition of some SiC to form Al-Si eutectic and Al_4C_3 .

Using travel speeds of 150 to 200 mm/min, alternating currents ranging from 145-160 amperes, and 100 % argon shielding, GTAW was conducted on the degassed samples. Welds exhibiting low porosity, good weld and base material fusion, and tensile properties equivalent to those produced in wrought alloys were reportedly obtained. However, the process was carried out with the addition of filler metal, and the presence of "small precipitates from the Al-SiC reaction" was noted.

Laser beam welding

Since the extent of the degradation of joint properties due to Al-SiC reaction has been shown to be a function of heat input in processes such as GTAW, welding processes in which the overall heat input can be minimized have been investigated as having good potential for use with respect to SiC/Al composites. One such welding technique which has been implemented in attempts to join these composites is laser beam welding (LBW), which is described as follows [48]:

In the LBW process, coalescence is produced by the heat from a concentrated, coherent (in phase), and monochromatic (singular frequency) beam of light, which is generated by stimulated electronic or molecular transitions to lower energy levels. As shown in Figure 14, the laser beam is directed by flat optical elements such as mirrors, and then focused on the surface to be welded by reflective focusing elements or lenses; focusing the beam at a small spot on the workpiece results in a high power density and low overall heat input. The process is generally carried out with the use of an inert shielding gas, in order to prevent oxidation of the molten puddle.

LB processes are based on the use of laser-emitting substances which produce electromagnetic radiation at specific frequencies (ultraviolet, visible, or infrared) when excited by various methods. Most lasers in use today are based on two of these substances. For industrial materials processing and welding, the 10.6 μm wavelength CO₂ laser, and the 1.06 μm wavelength YAG (neodymium-doped, yttrium aluminum garnet) laser are currently the predominant systems. The laser

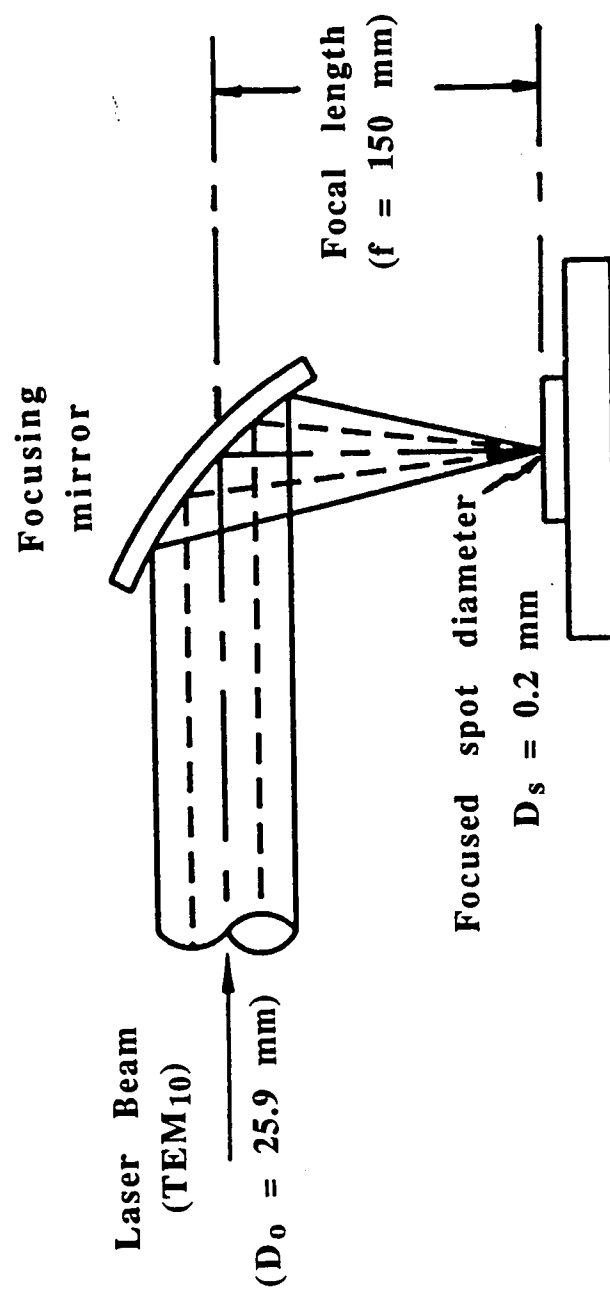


Figure 14. Schematic representation of LB focusing system.

Source: O'Brien, R.L., ed., *Welding Handbook*, Vol 2, 8th ed., Miami, Florida: American Welding Society, 1991, 74-107.

emitting substance in the CO₂ process is the CO₂ molecule, while in the YAG process it is the neodymium ion (Nd).

LB systems which utilize CO₂ are called gas lasers, and are currently the most efficient lasers available for high power material processing. In these systems, gas mixtures containing small amounts of CO₂ are subjected to an electrical discharge which excites the CO₂ molecule and causes it to emit laser light. Based upon the flow of the gas mixture, these lasers are categorized as slow axial, fast axial, or transverse. Output power levels ranging from 1 to 25 kW can be achieved using transverse flow, while the fast axial and slow axial flows provide levels of 500 to 6000 W and 50 to 1000 W, respectively. As opposed to gas lasers, considerably lower power outputs are possible in the solid-state YAG process. In this technique, the lasing material - a doped YAG crystal - is excited by means of an intense krypton or xenon lamp. The overall efficiency capability of this process is lower as compared to gas lasers, so the power outputs range from 10 to less than 1000 W.

The radial energy distribution of a laser beam is established by the so-called transverse electromagnetic (TEM) mode, which describes the cross-sectional shape of the beam. The TEM mode is expressed as TEM_{nm}, where the "n" and "m" refer to the beam's cross-section. The designation TEM₀₀ corresponds to the highest attainable quality and focusability, where the beam is characterized by a Gaussian power distribution. With sufficiently high beam power densities, deep penetration welding can be carried out as a result of an energy transfer mechanism called the keyhole welding phenomena. Energy inputs of the impinging laser beam become

so intense at high power density levels that they cannot be minimized by conduction, convection, or radiation. Therefore, melting and vaporization of the material at the impingement point of the beam occurs, creating a cavity or keyhole which is sustained by the vapor pressure of the material (Figure 15). Material which is molten but not vaporized forms a thin shell around this cavity, resulting in a narrow fusion zone. Under conditions such as these, depth-to-width ratios of approximately 10 to 1 can be obtained.

Thus, a narrow beam of light in LBW can be brought to an extremely sharp focus in order to produce an intense optical source when combined with a high initial power - generating temperatures that cause melting and vaporization of a material to occur within fractions of a second [51]. By virtue of these characteristics, accompanied by several other attributes shown as follows, LBW has been a candidate for use in joining of SiC/Al [52]:

- (1) regulated amounts of energy with a high power density can be delivered at low heat input, resulting in minimal thermal distortion and a narrow heat-affected zone.
- (2) the weld pool experiences rapid solidification, which produces weld metals having small cell sizes and improved properties.
- (3) no vacuum is required and x-rays are not generated, as compared to electron beam welding.

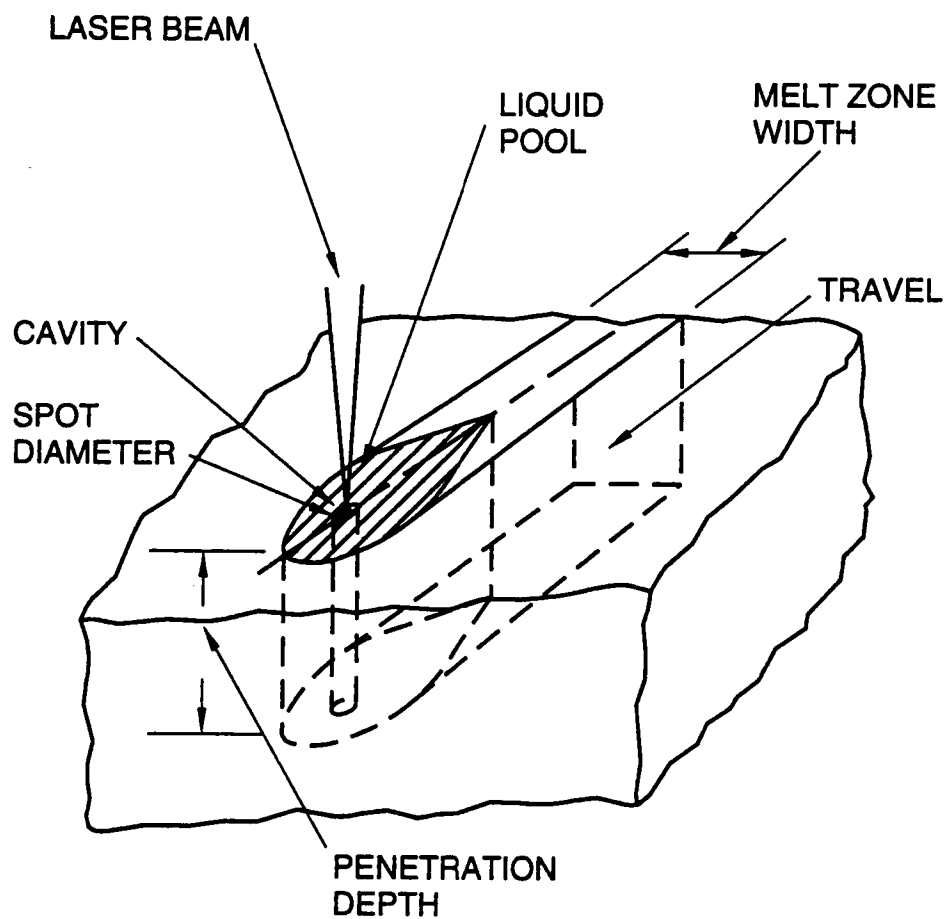


Figure 15. Schematic representation of keyhole welding.

Source: O'Brien, R.L., ed., *Welding Handbook*, Vol 2, 8th ed., Miami, Florida: American Welding Society, 1991, 74-107.

- (4) full automation and precise control can be achieved, as the laser beam may be shaped, directed, and focused with reflective optics.
- (5) shielding atmosphere can be freely selected under various pressures.

Problems in LBW of aluminum alloys include difficulty welding due to high thermal conductivity and high surface reflectivity, excessive drop-through of beads in full penetration welds, and mechanical property degradation. The degradation in mechanical properties is the result of porosity, the evaporation of strengthening elements in the weld metal during welding, and loss of strain-hardening structure in non-heat treatable alloys. In heat treatable alloys, strength reduction occurs due to the formation of solidification structure in the weld metal and the solid-solutionizing, annealing, and over-aging of precipitation hardening structure in the heat affected zone.[52]. In spite of these problems, the LBW process offers a precise control of welding parameters that is important to the successful application of fusion welding in relation to SiC-reinforced aluminum alloy composites.

Weldability studies of SiC/Al in relation to LBW

Dahotre, et al [53] carried out LB processing of SiC/Al composites with the objective of utilizing the focusing capability and the precise control over processing parameters in order to minimize the detrimental microstructural effects caused by Al-SiC reaction. In the investigation, bead-on-plate welding of 10 vol % SiC(p)-reinforced A356 was conducted utilizing a Rofin Sinar RS3000 laser. This instrument was a fast axial flow, 10.6 μm wavelength CO₂ laser having a power level capacity ranging from 0.2 to 3.7 kW.

During welding of the MMCs, the laser was operated in the TEM₁₀ mode using a pulsed beam, pure argon shielding, constant travel flow and beam diameter, and power levels of 1 to 2.5 kW. Rapid solidification welds were produced for different laser specific energies, and examination of the weldments showed that the extent of SiC/matrix reaction was a function of energy input. The high energy density associated with the LBW process resulted in the formation of aluminum carbide and the dissolution of SiC particulates for all heat inputs.

While the fusion zone consisted of a distribution of acicular aluminum carbide precipitates and blocky Si precipitates in a fine dendritic matrix, the size and amount of the plate-like precipitates decreased with decreasing specific energy. This variation of the volume fractions of Al₄C₃ ($V_{Al_4C_3}$) and SiC (V_{SiC}) as functions of specific energy was plotted, as shown in Figure 16. Thus it was suggested that under conditions of minimal superheating with short contact times between SiC and molten Al, the reaction could possibly be eliminated, which would be accomplished by operating the CO₂ laser in high-frequency, short duration pulse modes.

Microhardness profiles obtained from the base material through the fusion zone indicated high hardness values in the fusion zone, and a decrease in hardness occurring through the heat affected zone (HAZ) to the base material. A factor that was noted as a possible cause of high fusion zone hardness is the large difference between the thermal expansion coefficients of SiC and Al, the result of which is the generation of interfacial dislocations and the formation of precipitates along these dislocations. Higher HAZ hardness relative to that of the base material was attributed to the refined structure in the HAZ produced by a solid-state transformation.

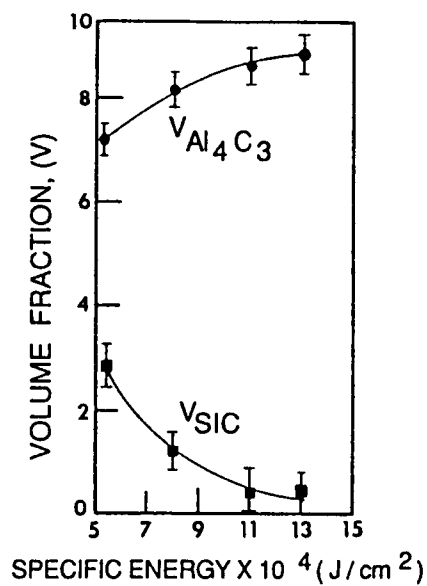


Figure 16. Volume fraction of aluminum carbide and SiC in laser-melted region as a function of specific energy.

Source: Dahotre, N.B., McCay, T.D., and McCay, M.H., Laser processing of a SiC/Al-alloy metal matrix composite, *Journal of Applied Physics* 65 (12), June 1989, 5072-5077.

An earlier study involving LBW of SiC/Al composites was carried out by Devletian [49], who evaluated this welding process in relation to 40 vol % SiC(p)/6061. Here, it was noted that the high energy-intensity laser beam produced a keyhole in which vaporization occurred, and that the resulting weldments were characterized by massive quantities of aluminum carbide and blocky Si.

Laser beam welding of 15 vol % SiC(p)-reinforced A356 was carried out by Cola, et al [54] using both a continuous wave CO₂ laser and a pulsed Nd:YAG laser in order to evaluate the effect of heat input and thermal cycle duration. Due to the low heat input levels (short thermal cycles) of laser welding, it was anticipated that SiC could be retained in the fusion zone with limited Al₄C₃ formation. A composite consisting of reinforced A356 was selected by virtue of the high Si content of the matrix alloy in order to reduce the extent of Al₄C₃ formation. Using the CO₂ laser, welds were produced at power levels of 1 and 1.5 kW and at various travel speeds in order to alter the heat inputs and produce thermal cycles of different durations. Welding by the Nd:YAG laser was conducted in the pulsed mode with an average power of 200 W and at various travel speeds. For both laser processes, helium shielding was employed.

Following LB processing, Cola, et al evaluated the composite specimens using optical microscopy, scanning electron microscopy, and energy dispersive x-ray analysis in order to characterize the microstructure from the base material through the weld. The results of this evaluation were as follows:

- (1) the base material comprised a uniform distribution of SiC particulates with coarse, primary alpha aluminum dendrites, and blocky Si in the inter-dendritic regions.
- (2) the fusion zone in CO₂ and Nd:YAG welds consisted of three distinct regions, (a,b & c):
 - (a) material adjacent to the base material had a distribution of SiC particulates (similar to the base material) surrounded by resolidified Al cells with Al-Si eutectic outlining the cell boundaries.
 - (b) an "intermediate" region was dominated by partially dissolved SiC particulates and acicular aluminum carbides, where many of the Al₄C₃ plates appeared to nucleate and grow from partially dissolved SiC. Here, the underlying microstructure consisted of primary Al cells and a greater amount of Al-Si at the cell boundaries relative to that in (a).
 - (c) in the center of the fusion zone, corresponding to the highest peak temperature, large Al₄C₃ plates, large blocky silicon nodules, primary Al cells, and Al-Si eutectic at cell boundaries comprised the micro-structure. Due to the lack of consistent orientation and the curvature of many of the carbide plates, it was suggested that the plates had nucleated and grown from the liquid. Si nodules appeared to have nucleated on the Al₄C₃ plates and grown with a spheroidal morphology into the liquid.

As in previous studies, the extent of SiC dissolution and carbide formation was dependent upon the heat input, where the shorter thermal cycles of the pulsed Nd:YAG process resulted in similar phase morphology as the continuous wave welds, although the phases were "finer" in appearance.

Cola, et al described the sequence of microstructural phase evolution that occurred during welding through the construction of a phase transformation model, which is shown in Figure 17. This model outlines the thermal cycles corresponding to the extremely rapid heating and cooling rates of LBW experienced in various portions of the welds, and correlates these thermal cycles to the resultant microstructural features. Corresponding to the thermal cycle "A" in Figure 17, the temperatures experienced by material adjacent to the keyhole (vapor cavity) are presumed to be near the vaporization temperature of Al of about 2500 °C. The resulting dissolution of 15 vol % SiC in liquid Al enriches the matrix alloy to about 7.5 % Si above its nominal composition of 6.5 to 7.5 % Si, creating a liquid hypereutectic at 15 % Si (the eutectic being at 12 %) that flows around the keyhole and fills in behind it as welding proceeds.

The temperature shown as T_{ceff} in Figure 17 is approximately 1950 °C, and is the point where Al_4C_3 precipitates from the liquid [38, 54], and subsequently grows with consumption of carbon from the liquid. As cooling continues, the temperature decreases to just below the liquidus of the hypereutectic composition (T_{leff}), whereupon Si precipitates from the liquid. Aluminum carbide plates serve as nucleation sites for the Si, which coarsen as cooling progresses. At T_{eff} - the effective eutectic temperature - remaining liquid solidifies as an invariant Al-Si cellular eutectic structure, appearing as Al cells with

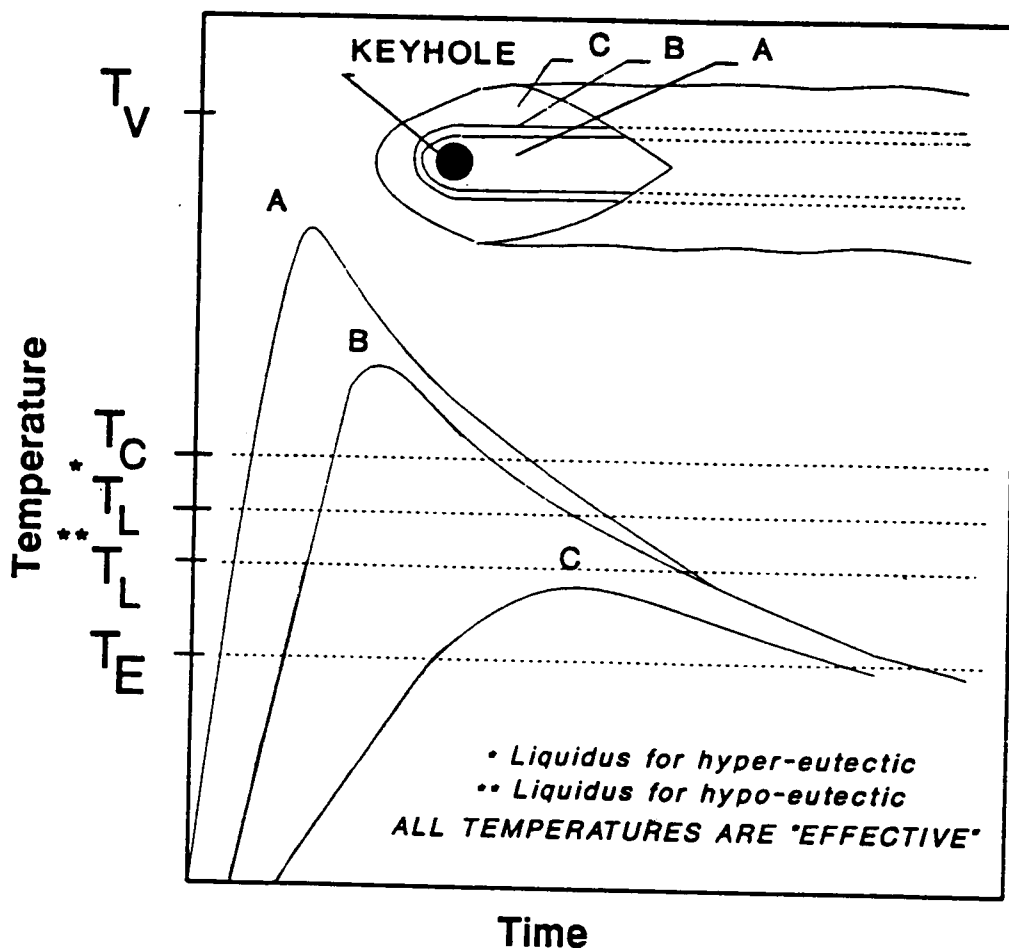


Figure 17. Qualitative model illustrating microstructural phase evolution for CO₂ LBW of SiC/Al [54].

Source: Cola, M.J., Lienert, T.J., Gould, J.E., Hurley, J.P., Laser welding of a SiC particulate reinforced aluminum metal matrix composite, *Conference Proceedings: Weldability of Materials*, Materials Park, Ohio: ASM International, 1990, 297-303.

intercellular Al-Si. In addition to this structure, the final microstructure consists of Al_4C_3 and solid Si.

In the portion of the fusion zone identified by thermal cycle "B" in Figure 17, the material is presumed to experience a peak temperature below that in "A", but sufficiently high to cause partial SiC dissolution and Al_4C_3 formation. Al-Si cellular eutectic structure forms from the partially Si-enriched remaining liquid. Although the peak temperature in the next portion of the fusion zone corresponding to thermal cycle "C" is above the liquidus of the hypoeutectic alloy composition, this temperature and heating rate are too low to promote SiC dissolution. Therefore, the solidified microstructure in this outer fusion zone consists of a hypoeutectic Al-Si cellular eutectic structure with a dispersion of SiC particulates.

Alternative joining techniques

Based on the results of investigations just described involving GTAW and LBW, it has not yet been possible to achieve full retention of the engineered properties of SiC-reinforced Al alloy composites using conventional autogenous fusion welding processes. However, alternative joining techniques with respect to these materials have been utilized with varying degrees of success. Devletian [49, 55], for example, developed a modified capacitor-discharge process for welding up to 40 vol % SiC(p)/6061 to itself or to an unreinforced aluminum alloy. The process was used with excellent results, producing weldments with no observable alteration of the SiC particulates and no evolution of porosity.

Capacitor discharge welding is a "stored energy" resistance welding process in which joining is accomplished by delivering a very short pulse with a high peak current (up to 300,000 amps in 30 milliseconds) to the pieces to be joined. The short high energy pulse can produce fine-grain diffusion-type welds with very little heat transfer beyond the weld zone [56]. In the investigation on SiC(p)-reinforced Al alloys using a modified capacitor discharge process [49], welds were produced between 6.4 mm (1/4") diameter composite rods and between rods and 3.2 to 6.4 mm thick sheets of 20% and 40% SiC(p)/6061. Rods were dropped from a height of about 76 mm, striking the faying surface of another rod or sheet. Upon impact, arcing and subsequent melting occurred by discharge of the capacitor bank. Data indicated that the arcing time was about 0.0004 seconds, and that more than half of the weld pool was expelled as spatter. Upon metallographic examination, the unetched microstructures of the fusion zone were free from porosity and virtually indistinguishable from those of the base material. The process was also used to join 48 % SiC(f)/6061 to unreinforced 6061, resulting in no apparent alteration of the continuous SiC fibers and a sound fusion zone and HAZ. In spite of these results, this capacitor discharge process has been applicable only for joining very small sections. The technique is limited to welding no more than 3/8 in. diameter composite studs or stud-to-plate welds [57].

Brazing using aluminum filler alloys has also been implemented as a method of joining SiC/Al alloy composites [2, 58, 59]. Morimoto, et al [59] investigated brazing of SiC(f)-reinforced composites, but reported degradation in properties when the brazing temperature was close to the melting temperature of the composite matrix due to the reaction between SiC and aluminum. As mentioned earlier, Ahearn, et al [50] used aluminum alloy filler metals in gas metal arc and GTA welding of 18 vol % SiC/6061, reportedly achieving

welds with low porosity and good fusion between the weld and base materials, although tensile strengths were in some cases lower than the optimum levels for these composites.

A design modification to improve the joining characteristics of cast SiC-reinforced Al alloys is an example of another method utilized to circumvent the problem of SiC/Al reaction. For electronic packaging materials such as those discussed earlier, joining must be carried out in order to create a hermetic seal; thus, the 20 % SiC/A357 electronic packages shown in Figure 18 have been fitted with unreinforced upper rims, upon which laser beam or electron beam welding can be subsequently performed [24].

The use of filler metals or welding of unreinforced surfaces may be suitable for certain applications where inconsistency of properties through the joint is acceptable. But where the full benefits of the engineered properties of the composite are desirable, the loss or non-uniform distribution of the reinforcing constituent can not be tolerated. In spite of current problems, fusion welding still offers the best potential means of preserving the integrity of SiC-reinforced aluminum alloys in a cost-effective manner. Therefore, a method of overcoming these problems is desirable to ensure the continued development of these composites.

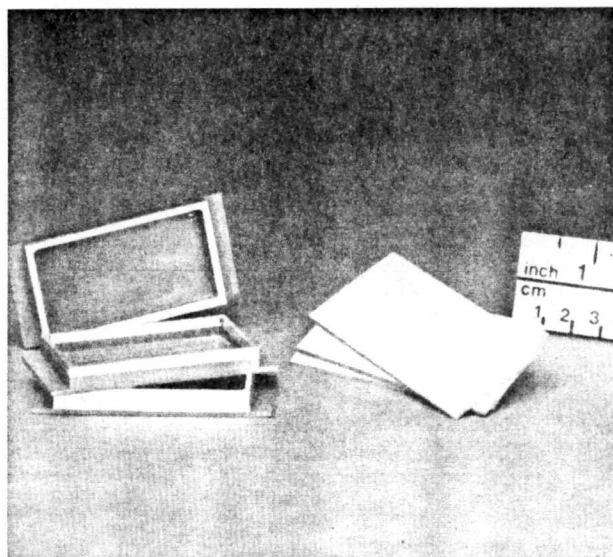


Figure 18. 65 volume % SiC(p)/A357 hermetic packages with lower reinforced sections [24].

Source: Kennerknecht, S., MMC studies via the investment casting process, *Conference Proceedings: Fabrication of Particulates Reinforced Metal Composites*, Materials Park, Ohio: ASM International, 1990, 87-100.

CHAPTER III

EXPERIMENTAL PROCEDURE

Material Employed

This investigation involved several aluminum alloy-matrix composites, where efforts were primarily focused on the evaluation of two silicon carbide particulate-reinforced alloys. The composite systems studied included SiC(p)/2024, SiC(p)/X2080, SiC(f)/1050, and graphite-reinforced 6061. Details regarding the as-received conditions and compositions of these materials are shown in Table 3.

In the preliminary phase of this investigation, weldability studies were carried out with respect to SiC(p)-reinforced 2024, supplemented by additional evaluation of the continuously-reinforced 1050 and 6061 alloys. Later stages of the study utilized the SiC(p)/X2080 composite.

Laser beam welding of SiC(p)/2024

LBW of SiC(p)-reinforced 2024 was carried out at the University of Tennessee Space Institute, Center for Laser Applications. The laser apparatus utilized was a 10.6 μm CO₂ Rofin-Sinar RS3000, which employs fast axial flow and can be operated in the continuous or pulsed mode with a maximum average power output of 3.7 kW. Prior to welding, the sample surfaces were prepared by brushing with stainless steel wire and cleaning with acetone. The purpose of wire brushing was to increase the surface

TABLE 3
Composite systems employed

Matrix alloy composition (wt % alloying element)	Type of reinforcement	As-received condition	Supplier
2024 4.4 Cu, 0.6 Mn, 1.5 Mg, balance Al [45]	15 volume % SiC particulates	0.7, 1.2, and 1.6 cm thick plates. T3 temper.	Alcoa
X2080* 3.8 Cu, 1.8 Mg, 0.2 Zr, balance Al	20 volume % SiC particulates	0.2 cm-thick, hot-rolled sheet. As-fabricated. (F-temper)	Alcoa
1050 99.50 Al	40 volume % continuous, unidirectional SiC "Nicalon" fibers	0.2 cm-thick roll diffusion bonded sheet	Fuji Heavy Industries, LTD.
6061 0.6 Si, 0.28 Mn, 1.0 Mg, 0.20 Cr, balance Al [45]	unidirectional graphite fibers	0.2 cm-thick sheet	University of Tennessee

(* denotes an experimental alloy)

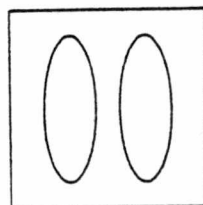
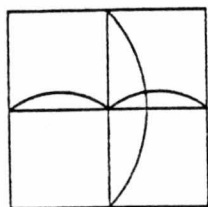
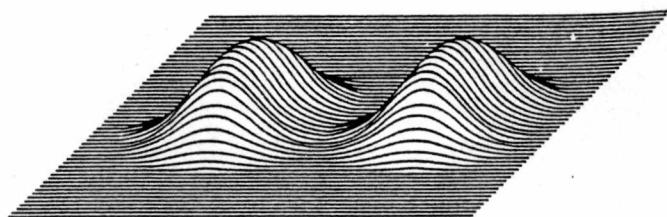
roughness of the specimens in order to reduce reflectivity and improve energy absorption, as well as to provide cleaning. Immediately following the sample preparation, bead-on-plate LB welding was conducted in the continuous wave mode on 0.7 and 1.2 cm-thick 15 vol % SiC(p)-reinforced 2024 plates. The beads were produced using pure argon shielding, a beam focal point situated just below the plate surfaces, and a power level of 3 kW corresponding to a power density exceeding about 10^5 Watts/mm². In order to observe the effect of heat input, travel speed was varied from 1.5 to 4m/min. A summary of the welding parameters is given in Table 4.

Figure 14 shows a schematic representation of the LBW focusing system with respect to some of the specific parameters given in Table 4. The spatial mode utilized was the TEM₁₀ mode, where the beam cross section corresponding to this mode is shown in Figure 19 [48]. Using a parabolic focusing mirror of 150 mm in focal length, the approximately 26 mm diameter laser beam was focused to a spot size of about 0.2 mm. Since the focal point of the beam was set below the surface of the plates, the defocused distance, f_d , was written as - 1mm.

Bead-on-plate LB welding was conducted on the 0.7 cm-thick SiC(p)/2024 plate in two perpendicular directions. The surface of the specimen designated “A “ was cleaned with a stainless steel wire brush and washed with acetone prior to welding. For both beads, a travel speed of 1.5 m/min was used. On the 1.2 cm-thick SiC(p)/2024 plates, continuous wave LBW was also conducted such that beads were produced in two directions. Prior to welding, the surfaces of these samples were scrubbed with 600 grit sandpaper and washed with acetone. Travel speeds of 1,2, and 4 m/min were used to weld

TABLE 4
CO₂ LBW Parameters
(Maximum power: 3.7 kW)

Operating mode:	continuous wave (cw)
Laser power:	3 kW
Travel speed:	1.5 - 4 m/min
Coaxial and/or shielding gas flow rate:	10 liters/min argon
Beam spatial mode (transverse electromagnetic mode):	TEM ₁₀
Focal length, f:	150 mm
Defocused distance, f_d:	- 1mm



TEM₁₀

Beam Energy Density Beam Cross-Section

Figure 19. Beam cross-section and energy density corresponding to the TEM₁₀ mode.

Source: O'Brien, R.L., ed., *Welding Handbook*, Vol 2, 8th ed., Miami, Florida: American Welding Society, 1991, 74-107.

a 1.2 cm-thick specimen (sample “B”), while 1 and 2 m/min speeds were utilized with respect to another sample (“C”) of the same thickness.

Two shielding methods were utilized during LB welding. In the first of these methods, only the coaxial gas flow through the nozzle provided shielding. A second method included shielding gas in addition to the coaxial gas flow, while a third method (coaxial gas + plasma control gas) was not implemented based on a previous study indicating its possible negative effects in relation to LBW of Al alloys [60]. In summary, the conditions discussed above corresponding to LBW of the three specimens are given in Table 5.

LBW of SiC(f)/1050

Bead-on-plate LB welding was conducted on a 0.2 cm thick sheet of 1050 (99.5 % aluminum) reinforced with 40 volume % continuous SiC fibers. The welding was performed at XDE Systems, Inc. of Sunnyvale, California, utilizing both a CO₂ laser and an Nd:YAG laser. Using the CO₂ process, beads were produced in the transverse direction (at right angles to the orientation of the reinforcing SiC fibers) while operating the laser in the continuous wave mode at a power level of 500 W. LB welding was conducted at travel speeds varying from about 0.25 m/min to 1.3 m/min, in order to observe the effect of heat input.

Operating the Nd:YAG laser in the pulsed mode, bead-on-plate welding was conducted in the longitudinal direction - that is, parallel to the direction of fiber orientation. An average power level of 200 W was delivered at a rate of 10 pulses/second, and

TABLE 5
LBW conditions for SiC(p)/2024 plate specimens

Specimen	Travel speed	Argon flow rate
A, 0.7 cm thick plate	1.5 m/min	coaxial flow of 10 liters/min
B, 1.2 cm thick plate	1 m/min (first bead), 2 m/min and 4 m/min at 90° to first bead.	coaxial flow of 10 liters/min
C, 1.2 cm thick plate	1 m/min (first bead), 2 m/min at 90° to first bead.	coaxial flow of 10 liters/min with shielding gas addition of 10 liters/min.

shielding was provided by nitrogen. Travel speed in this case ranged from 0.05 m/min to 0.41 m/min. The welding parameters for both CO₂ and Nd:YAG processing are summarized in Table 6.

GTAW of SiC(p) and graphite-reinforced Al alloys

SiC(p)-reinforced 2024 was evaluated with respect to GTAW with the objective of comparing the tendency for SiC dissolution and aluminum carbide formation in this process relative to that in high energy density LB processing. Using GTAW with helium shielding, standing edge welding was conducted on 1/8" thick 15 volume % SiC(p)/2024 plates. The heat input was altered during welding by varying the current level along the length of the joint, while maintaining a constant travel speed. For the 20 volume % SiC(p)/X2080 composite, this procedure was repeated so that standing edge welds could be produced on 1/16" thick specimens using currents ranging from 115 to 140 amperes. A schematic representation of the joint configuration for both materials is shown in Figure 20, and a compilation of the GTAW parameters is shown in Table 7.

On a 0.2 cm thick sheet of graphite-reinforced 6061, GTA spot welding was performed in order to determine the effect of high heat input on the resulting microstructure and properties. The sample was subjected to spot welding for a period of about 5 seconds, while utilizing helium shielding. For a SiC(p)/2024 plate having a thickness of 0.7 cm, this procedure was repeated.

TABLE 6
LBW parameters for processing SiC(f)/1050

	Laser	
	CO₂	Nd:YAG
Power:	500 W	200 W pulse rate: 10 pps
Shielding gas:	nitrogen	nitrogen
Travel speed, in/min (m/min):	10 (0.254)	2 (0.051)
	25 (0.635)	4 (0.102)
	50 (1.27)	8 (0.203)
		16 (0.406)

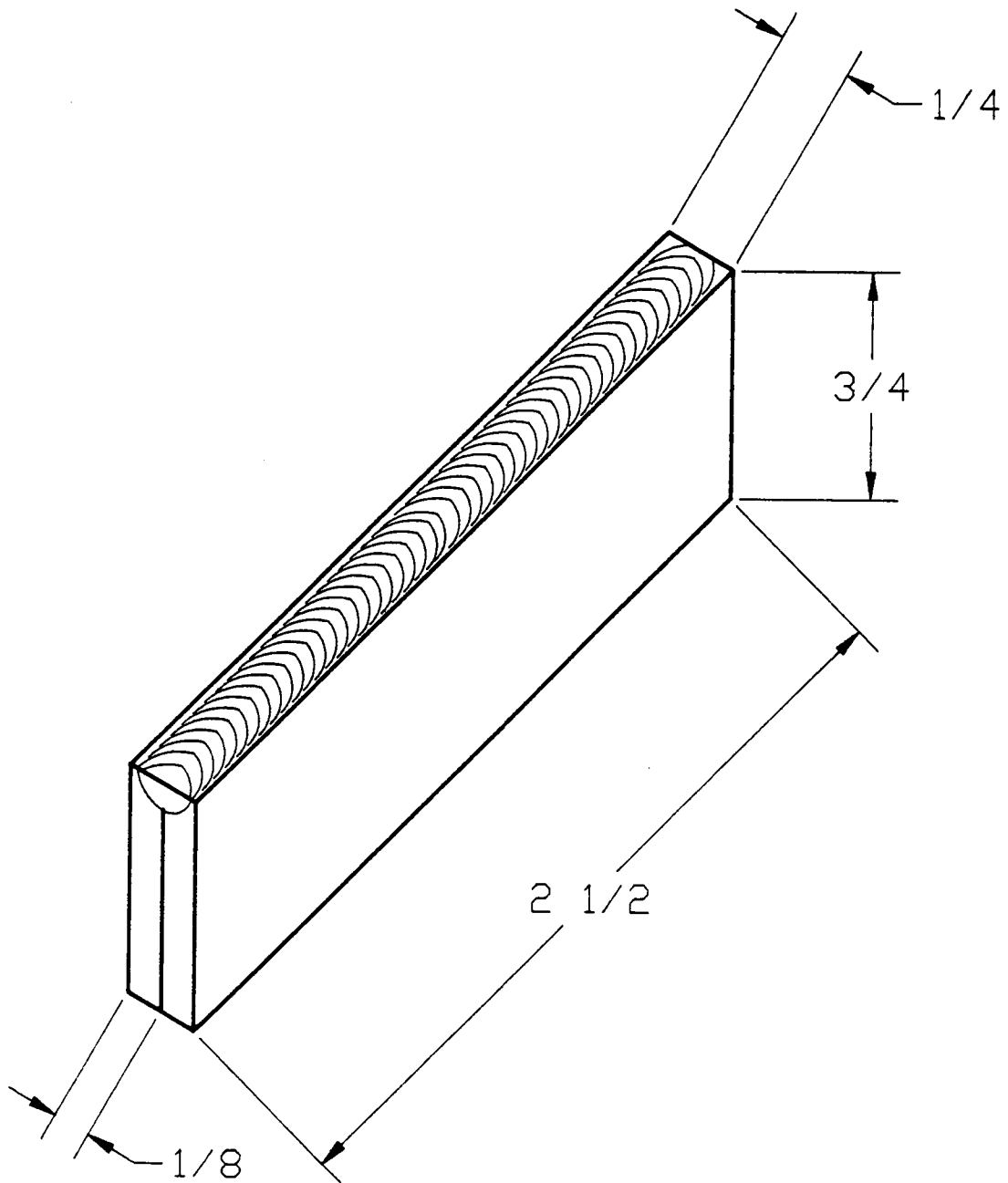


Figure 20. Schematic representation of joint configuration for standing edge welding of SiC(p)-reinforced 2024 and X2080.

TABLE 7
GTAW parameters for standing edge welds in
SiC(p)/2024 and SiC(p)/X2080

Material:	15 vol % SiC(p)/2024	20 vol % SiC(p)/2080
Plate dimensions: (length x width x thickness)	~8.9 x 1.6 x 0.32cm (3 1/2 x 5/8 x 1/8in)	5.72 x 1.59 x 0.159cm (2 1/4 x 5/8 x 1/16in)
Current:	DCEP 115, 170, and 200 A	DCEN 115 to 140 A
Travel speed:	0.152 m/min (6 in/min)	0.152 m/min (6 in/min)
Electrode:	W-ThO ₂ 5/32" diameter (0.4 cm)	W-ThO ₂ 1/8" diameter (0.318 cm)
Shielding gas:	0.71 m ³ /min helium (25 ft ³ /min)	0.85 m ³ /min helium (30 ft ³ /min)

Microstructural Examination of SiC/Al composites

Specimens were prepared for metallographic inspection by optical light microscopy (OLM), following LBW and GTAW. The orthogonally-situated beads produced by LBW on the 0.7 cm thick SiC(p)/2024 plate were sectioned so that microstructural profiles in the XY, YZ, and XZ directions could be examined. For all other specimens, sectioning was carried out to obtain transverse profiles of the LB and GTA weldments. Samples were then mounted in epoxy, and were polished using up to 1 μm diamond paste. Subsequent etching of the 2024, X2080, and 6061-based composites was completed utilizing a modified Keller's reagent consisting of 2.5 ml HNO_3 , 1 ml HCl , 1.5 ml HF , and 95 ml H_2O . Normally, etching of these specimens was followed by a dip in concentrated HNO_3 in order to remove "smut" that formed as a result of the samples' immersion in the etchant. The SiC(f)-reinforced 1050 samples were etched with a reagent suited for higher purity aluminum alloys, which contained 100 ml HCl , 100 ml HNO_3 , 100 ml H_2O , and 25 ml HF .

Following initial evaluation of the composites by OLM, scanning electron microscopy (SEM) and energy dispersive spectroscopy (EDS) were employed in an effort to characterize the microstructural features in some of the specimens. SEM was also utilized to examine fracture surfaces obtained from LB and GTA welds. While specimens for fractographic inspection were produced during tensile testing of LB welds in SiC(p)/2024, GTAW fracture surfaces were obtained by separating the SiC(p)/2024 plates which had been joined by standing edge welding.

Aging of welded SiC(p)/2024 specimens

Aging studies were conducted on SiC(p)/2024 specimens subjected to LB processing without supplemental shielding gas, in order to determine the effect of post weld heat treatment (PWHT) with respect to the metallurgical response in the fusion zone. Prior to heat treating, specimens containing portions of beads produced at 1 and 2 m/min travel speeds were sealed in evacuated quartz tubes. Aging was then performed at 300 °F (149 °C) and 400 °F (204 °C) for 24 and 72 hour periods, followed by air cooling of the samples within the tubes.

Mechanical properties

Tensile specimens were machined from a 0.7 cm thick SiC(p)/2024 plate containing LB welds produced at a travel speed of 1.5 m/min. As shown in Figure 21, flat, dogbone specimens having gauge lengths of approximately 1/2" (1.27 cm) and cross-sectional areas of about $2 \times 10^{-3} \text{ in}^2$ (0.013 cm^2) were cut by EDM, such that the center of the specimens either contained welds corresponding to the longitudinal (rolling) direction of the plate or welds oriented normal to this direction.

In order to reveal the fusion zone, the tensile specimens were etched with a concentrated Keller's macroetchant prior to testing. Subsequent tensile testing using an Instron tester was performed at a crosshead speed of 0.01 in/min (0.0254 cm/min) for all samples. Following testing, fractured specimens were subjected to fractographic inspection by SEM.

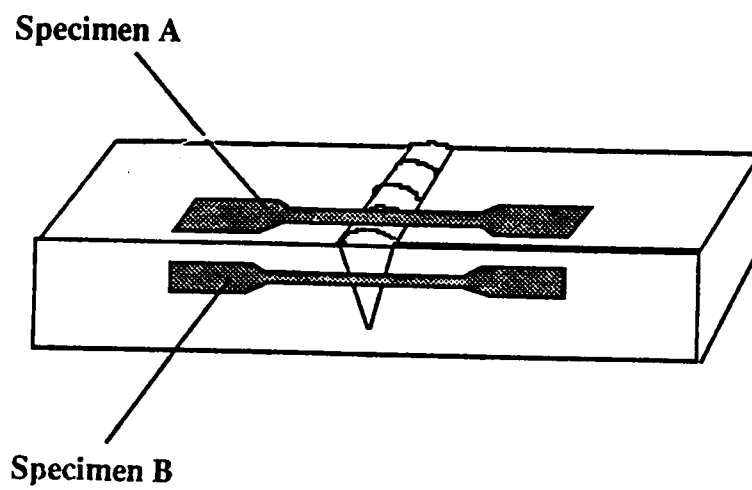


Figure 21. Locations of tensile specimen extraction in LB-welded SiC(p)/2024 plate.

Evaluation of mechanical properties of LBW and GTAW specimens was also carried out by obtaining microhardness profiles across the weldments. Vickers hardness evaluations with a 1 kg load were conducted on SiC(p)/2024 specimens in the as-welded and heat treated conditions, and on the specimen subjected to manual GTAW.

Based on the microstructural features and mechanical properties of the fusion zone in LB and GTA welds, tests were conducted to determine the susceptibility of the weldments to degradation in an aqueous environment. A SiC(p)/2024 sample processed by LBW was submerged in distilled water for a period of several days, after which the appearance and mechanical properties of the weldment were evaluated. The fusion zone of spot-welded graphite-reinforced 6061 was subjected to air exposure, and was similarly examined.

Construction of MMC welding apparatus

The application of fusion welding processes to SiC/Al results in degradation of the engineered properties of the composite due to matrix-reinforcement reaction, as described in chapter 2. Although this reaction can not be completely inhibited by altering the welding parameters, studies such as those carried out by Cola, et al [54] show that the reaction can be limited to portions of the weld associated with specific thermal cycles. Cola, et al constructed a model illustrating microstructural phase evolution based on CO₂ LBW of 15 vol % SiC(p)/A356 (Figure 17, chapter 2), noting that aluminum carbide forms readily in the keyhole region in which temperatures are presumed to approach the vaporization temperature of Al of about 2500 °C. In the adjacent "intermediate" zone, partial SiC dissolution and Al₄C₃ formation are attributed to peak temperatures that exceed about 1950

°C, but are lower than those encountered in the keyhole region. An absence of Al_4C_3 particles and a base material distribution of SiC particulates characterizes the so-called "outer" fusion zone; here, peak temperatures are insufficient to cause SiC dissolution, although they are higher than the liquidus of the matrix alloy.

In an effort to produce a weldment characterized primarily by this "outer" fusion zone microstructure, the apparatus shown in Figures 22 and 23 was constructed to "mechanically" alter the weld puddle during fusion welding of SiC/Al. The desired function of the apparatus is to upset two composite coupons during GTA butt or GTA spot welding, such that the portion of the weld puddle that experiences temperatures responsible for matrix-reinforcement reactions will be expelled. Thus, with only the "outer" fusion zone remaining to solidify and affect the bond between the coupons, the base metal distribution of SiC will be effectively retained throughout the resultant bond region.

The primary components of the apparatus include a motor-driven slide assembly upon which the coupons are mounted, and a pair of clevis-headed pneumatic cylinders between which the slide assembly travels. Squeeze rollers are pinned within each clevis, which bear upon the outer edges of the coupons upon actuation of the cylinders and provide a force to squeeze the coupons across the bond region. Additional components - including cylinder controls and a gearmotor - are mounted along with the slide and cylinders on a light-weight base plate. This portable apparatus can be potentially used with various fusion welding processes, where the slide assembly would travel under a stationary electrode or laser beam during butt welding or remain fixed for spot welding. Features which comprise the mechanical assembly shown in Figures 22 and 23 are described as follows:

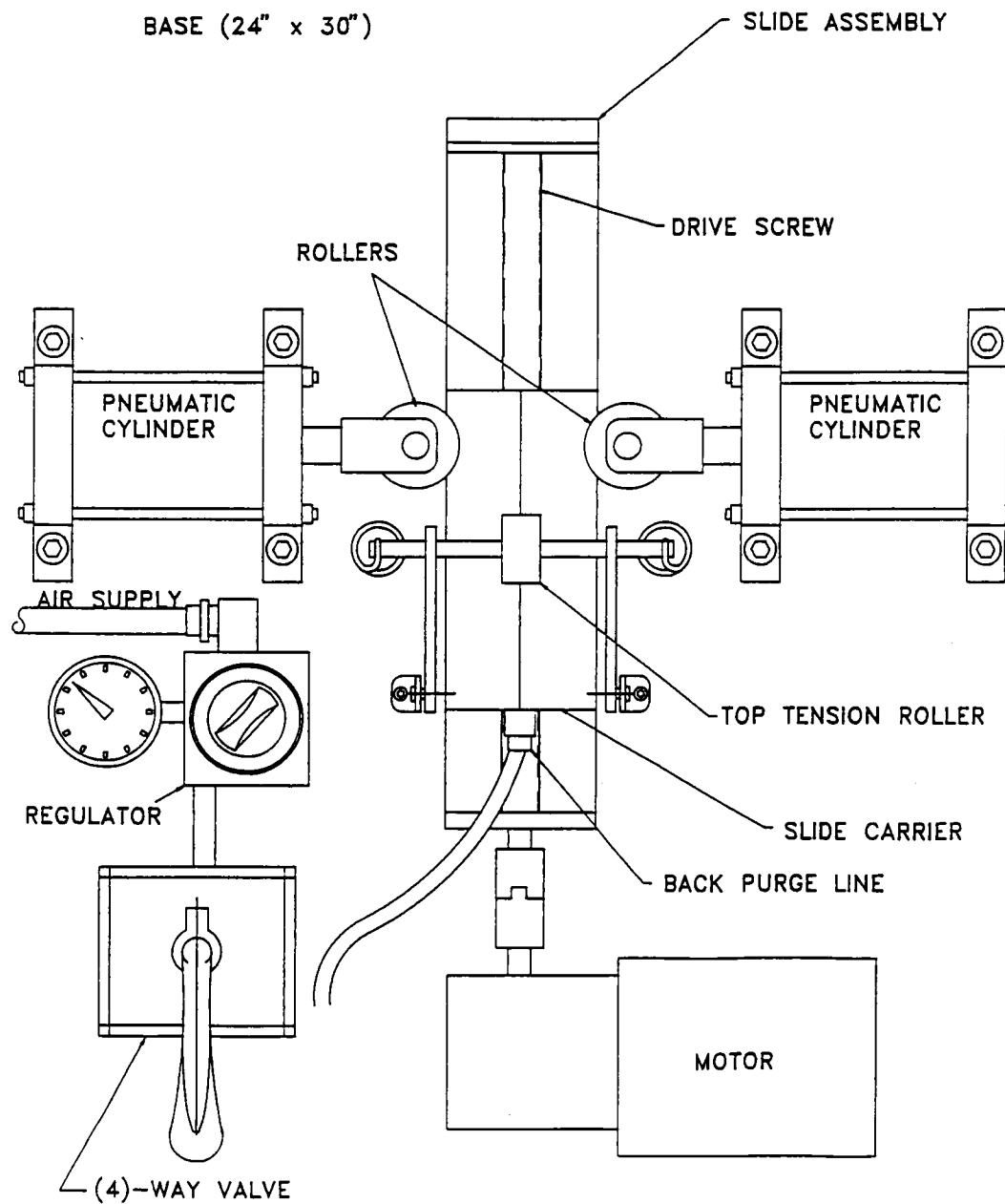


Figure 22. MMC welding assembly.

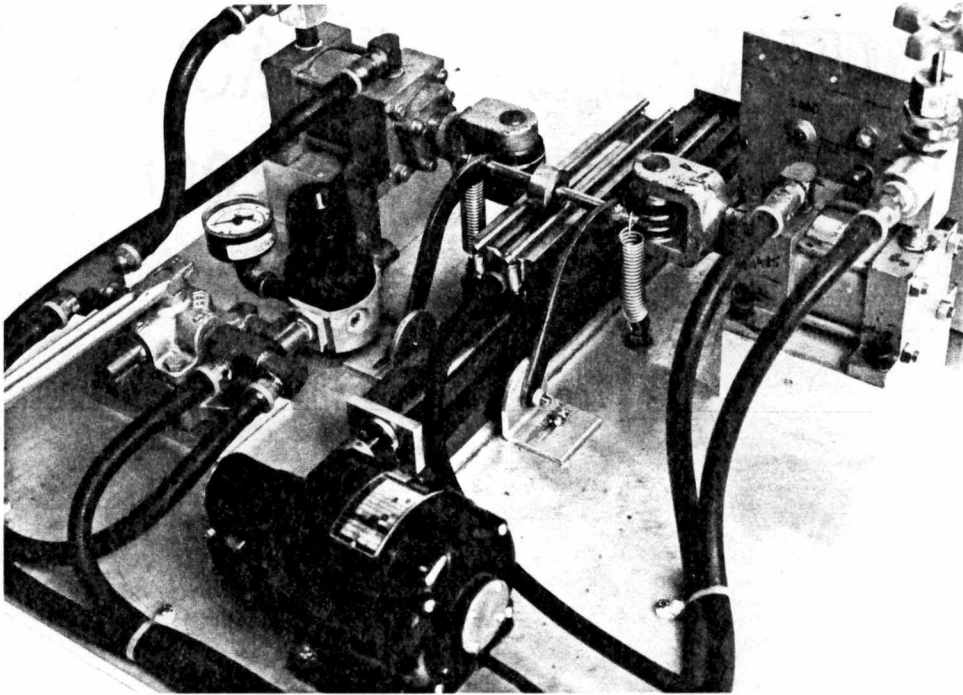


Figure 23. MMC welding assembly

- (a) mounting block - the specimen or specimens to be welded are mounted on a hollow copper chill block in order to optimize conduction of heat away from the samples during the welding operation. Having a length of about 8 inches and width of 1 1/2", the block can accommodate samples of up to the same length, while it is necessary for the total width of the samples to exceed the block width by at least 1/4 inch. To carry out back purging (provide under-bead shielding) during welding, the block incorporates a 1/2" wide, u-shaped, shallow trough, which is centered along the block surface and contains several 1/32 " diameter gas inlets.
- (b) hold-down straps - in order to situate the specimens to be welded on the copper mounting block, a pair of 1/4" diameter stainless steel "straps" are positioned along the length of the block and attached to it via a set of four linkage springs. The 1/4" diameter springs link each grooved strap-end to a screw bolted into the base of the mounting block. Along with the purpose of inhibiting the tendency for vertical movement of two specimens subjected to horizontal loading, the spring-loaded strap design is utilized for ease of specimen placement and removal. This design also provides for the use of samples of varying dimensions without the need for machining in order to attach the specimens to the mounting block. Finally, the utilization of springs provides the ability to accommodate overloading or thermal expansion of samples without causing damage to the apparatus that may result from a more rigid design.

- (c) end plate - because specimens are subjected to lateral forces while being transported between two drive rollers under the welding apparatus, a thin, aluminum plate is placed on one end of the copper mounting block in order to prevent the specimens from sliding off the end of the block when pressure is applied. This "end plate" is positioned such that it juts above the mounting block by 1/16", thereby inhibiting motion of specimens in one direction.
- (d) slide assembly - the copper mounting block, upon which the specimens to be welded are placed, is bolted to the carrier block of a motor-driven slide assembly. Consisting of a heavy duty recessed base machine slide driven on a 1/2 - 13 Vee type lead screw, the assembly provides 10 inches of travel with its 8" long carrier block sliding on an 18" long base. The lead screw is supported by self-lubricating bearings at either end of the slide base, and is fitted with a coupling on one end for direct connection to the motor.
- (e) gearmotor - the slide assembly is driven by a 1/20 HP, 115 V DC shunt-wound reversible gearmotor. Initially, a right angle gear reducer was mounted onto the motor, providing a 10:1 gear ratio for a maximum speed of 170 rpm and a torque rating of 13 in-lbs. This gear reducer was replaced by another in subsequent testing in order to increase the travel speed capability of the slide assembly. With a new gear ratio of 5:1, the speed was increased to 345 rpm and the torque rating was reduced to 6 in-lbs, thereby resulting in a maximum slide travel speed of about 20 in/min.

- (f) "squeeze" rolls - In order to "mechanically" affect the solidification characteristics of the weld pool, a lateral force is applied to a point on the outer edge of each of the abutting specimens during welding by means of two squeeze rolls situated on either side of the slide assembly. The rollers utilized are 1 3/4" diameter filler-metal feed rollers, which consist of an insulating, epoxy-based material encircled with a grooved and notched steel band for bearing on the edges of the specimens. In order to evaluate the effect of upset location on the weld pool behavior, the point of contact between the rollers and the specimens relative to the position of the welding apparatus (the electrode, for example) can be varied.
- (g) clevis-headed pneumatic cylinders - The force applied by the drive rollers during butt welding is controlled by the actuation of two intermediate pressure, square head, non-rotating pneumatic cylinders. Having threaded piston rod ends, the cylinder heads are fitted with female clevises within which the drive rollers are pinned. The cylinders are situated relative to the slide assembly in order to achieve a 1/8" clearance between the slide carrier and the clevis. Additionally, they are positioned facing each other (on either side of the slide assembly) such that their alignment with respect to each other's point of contact with the specimen edges is within 0.001 inch. With the cylinder 2 1/2" bore diameter and 1" rod diameter, the force corresponding to a forward stroke and a fluid (air) pressure of 100 psi is given as approximately 491 lbs; the stroke length is 1 inch. The inlet portholes of both cylinders are fitted with needle valves for equalizing the

stroke timing, which is otherwise inconsistent between the cylinders due to differences in the lengths of tubing connecting the portholes to the control valve.

- (h) directional control valve and pressure regulator - a four-way, three-position hand valve having a maximum pressure rating of 250 psi is used to actuate the double-acting pneumatic cylinders. The air pressure is controlled using a standard regulator with a maximum pressure rating of 300 psi.
- (i) base plate - the above components are mounted on a flat, 1/2" thick, aluminum alloy base plate having dimensions of 27 3/4" long by 2 feet wide.
- (j) top tension roller - in order to inhibit the tendency for buckling of the samples to occur upon actuation of the cylinders, it is necessary to supplement the action of the hold-down straps discussed in (b). A device incorporating a spring-loaded roller was constructed, which provides a significant downward force on the specimens that reduces the occurrence of "cross-over" or buckling of the samples. This is accomplished by means of linking a cylindrical rod, upon which a steel roller is placed, to a slotted bracket mounted on the aluminum base plate. Supporting a 5/8" thick roller that is inserted between the hold-down straps, the rod is fitted with two linkage pins on either side of the roller, and two linkage springs on its ends. As the pins are mounted on the base plate via a slotted aluminum bracket,

the position of the pins, and hence that of the roller on the specimens, is adjustable. The overhead load on the specimens can also be varied by utilizing different linkage springs, which are hooked onto the grooved ends of the threaded rod, and to eyebolts mounted on the base plate. Using music wire, stainless steel, and zinc-plated steel springs, load ratings ranging from about 3 lb/in to 34 lb/in can be utilized.

Preliminary testing

Initial testing of the MMC apparatus was conducted using an Airco DC GTA welder. Rather than utilizing SiC/Al composite material in these preliminary tests, stainless steel test specimens were used in order to evaluate the assembly mechanism and to establish initial welding parameters and a suitable specimen geometry. The samples tested were approximately 1/16" thick 304L stainless steel coupons, having lengths not exceeding the length of the copper mounting block and widths slightly greater than that of the block. A set of single, 2 inch-wide strips were initially tested in order adjust the cylinder actuation and check the effectiveness of the method by which the samples were fixed on the mounting block. After the addition of needle valves to the air cylinders and an end plate to the mounting block, bead-on-plate welding was conducted on the test coupons.

Samples utilized in preliminary tests to determine the feasibility of mechanically eliminating a portion of the fusion zone during welding were 304L stainless steel test coupons about 1/16" thick and up to 8 inches long. In order to establish initial parameters before actuating the cylinders during welding, bead-on-plate welding was conducted on 2 inch-wide samples, and butt welding was carried out to join 1 inch-wide coupons. The

welding parameters included current level, travel speed, shielding gas flow rate, back purging flow rate, and arc length. Adjustments to these individual variables were made in order to achieve the desired weldment characteristics. Upon determination of the initial parameters, subsequent testing was conducted to evaluate the effect of applied pressure on the weld pool during butt-welding. This testing involved actuation of the pneumatic cylinders with the objective of expelling a portion of the weld pool. In addition to the aforementioned welding parameters, variables in this procedure included the position of the rollers with respect to the weld pool, the amount of pressure supplied to the air cylinders, the loading rate of the top tension roller, and the specimen geometry.

Coupons utilized in this series of tests were machined such that a portion of the faying surface was tapered to a specific degree. The purpose of the taper was to maintain a gap between the abutting edges of the coupons ahead of the weld pool, where this gap would be wide enough to allow gradual advancement of the rollers as welding progressed. Like the coupon shown in Figure 24, most test specimens were machined with taper angles ranging from about 1 to 3 degrees. The untapered portion of the samples varied from about 1/2" to 1 1/2" in length.

After placing the test coupons under the spring-loaded hold-down straps as shown in Figure 25, the coupons were then centered on the slide assembly such that their abutting edges were positioned directly under the stationary electrode. The base plate was aligned so that this position would be maintained as the slide carrier advanced. Next, the location of the electrode with respect to the "squeeze" rolls was selected, where the electrode was placed either in front of or behind the rollers at various distances, or directly in line with them. After adjusting the regulator to the desired air pressure level, the load rating on the

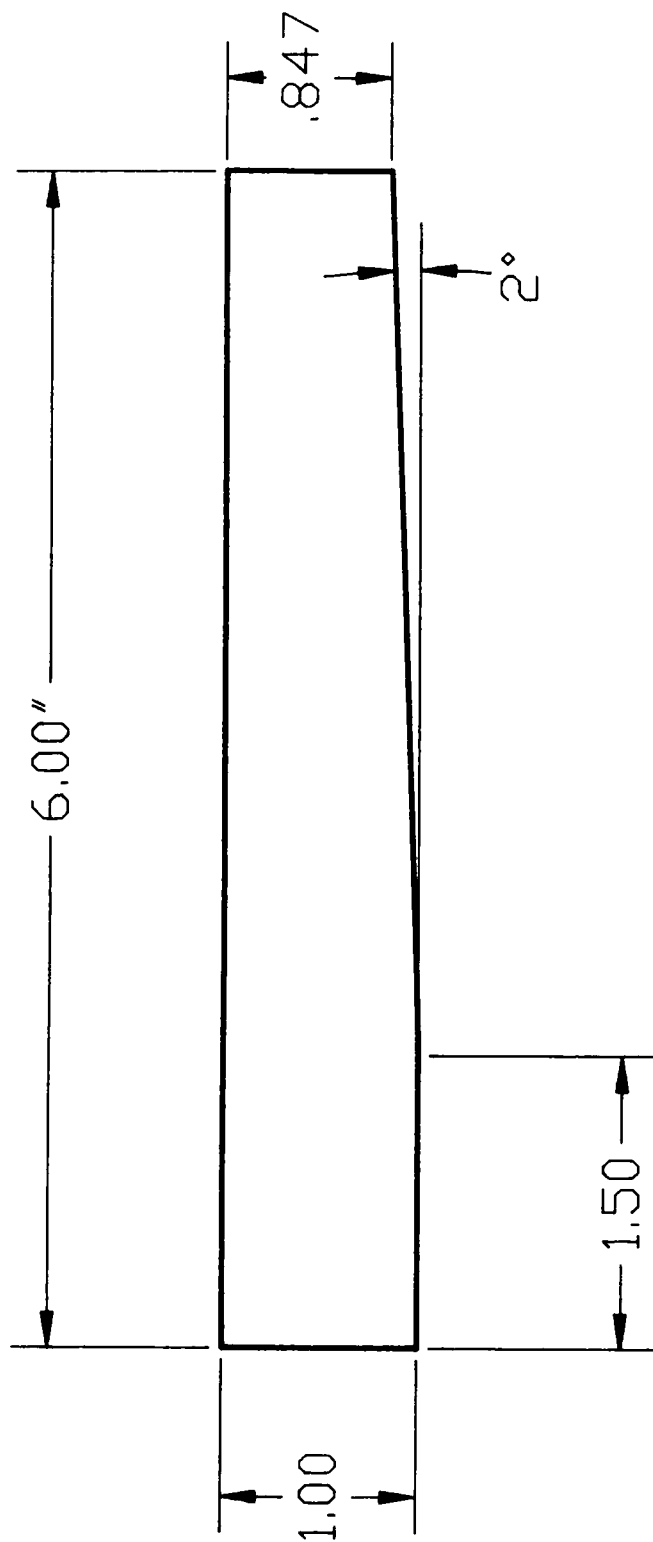


Figure 24. Schematic representation of typical test coupon for GTA butt welding.

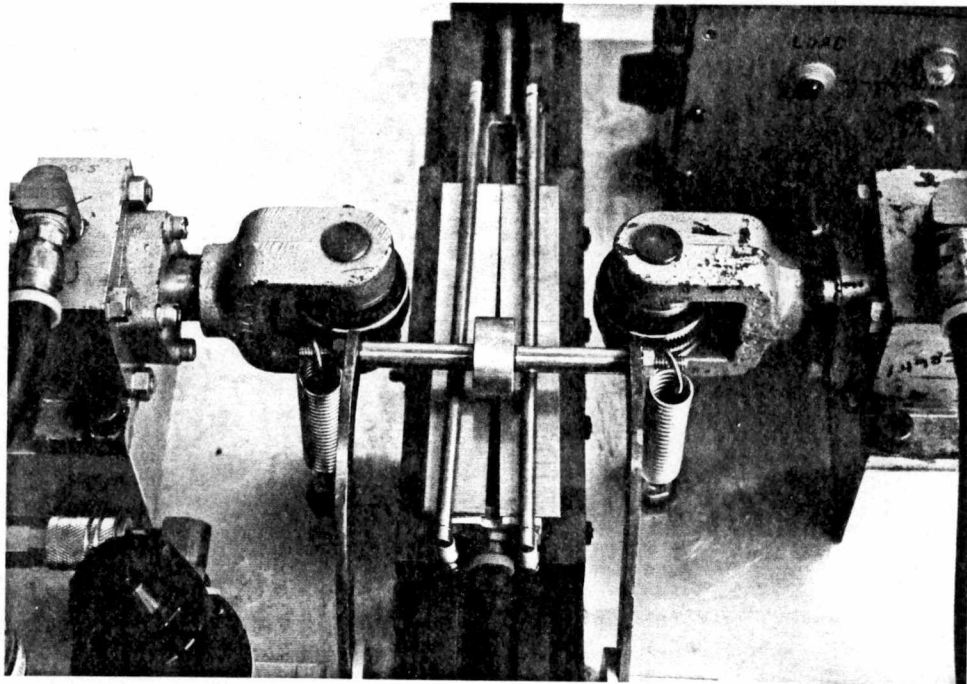


Figure 25. Set-up for GTA butt welding of stainless steel test coupons.

top tension roller was set (in some of the tests) by attaching an appropriate pair of springs to the tension roller rod. The welding process was then initiated, where the slide carrier was driven only after the weld pool had been established. When the weld length just reached or exceeded that of the untapered section of the test coupons, the pneumatic cylinders were actuated. The corresponding advancement of the rollers was very rapid, such that the specimens were fully loaded almost instantaneously. The initial level of pressure was maintained as welding progressed.

In addition to the stainless steel coupons, aluminum alloys were utilized in preliminary testing of the MMC welding device. Specimens comprised of alloys 2024 and 3003 were machined having geometries similar to those of the 304L coupons, and the test procedure utilized with the steels was repeated for these aluminum alloys. In order to provide the cathodic cleaning action important for welding aluminum, equipment which could be operated with alternating current was required. A Hobart GTA welder was employed for this second series of tests. Welding parameters for both the stainless steel and aluminum alloy coupons are summarized in Table 8.

GTA spot welding

The effectiveness of the MMC welding device was also evaluated with respect to GTA spot welding, where pressure was applied to two 1/16"-thick triangular-shaped 2024 aluminum alloy coupons during welding. The coupons had one chamfered corner with a length ranging from about 1/4" to 5/16". The specimens were mounted on the copper chill block such that their chamfered corners were abutted and centered under the electrode. Each coupon was approximately 1 inch-wide, while the length of each side opposite to the

TABLE 8
Parameters for evaluating MMC welding fixture
using stainless steel and Al alloy test coupons

Welding parameters	304L	2024, 3003
Welding apparatus	Airco DC arc welder	Hobart AC/DC arc welder
Specimen geometry	1/16 " thick, 5 - 8 " length (including 1/2 to 1 1/2" untapered), 1 to 4 degree taper	1/16" thick, 5 - 8" length (including 1 to 1 1/2" untapered), 1 to 2 degree taper
Specimen preparation	clean faying surfaces with stainless steel brush	clean entire specimen with acetone, scrub with an abrasive, file faying surfaces
Current	DCEN, 80 to 95 amperes	AC, 70 - 105 amperes
Electrode diameter	1/16"	1/8"
Arc length	1/16"	1/32 to 1/16"
Argon flow rates	30 CFH shielding and back purging	30 CFH shielding and back purging
Travel speed	11 - 21 in/min	7 - 15 in/min
Load rating of top tension roller springs	3 - 34 lb/in	3 - 7 lb/in
Regulated air pressure	30 - 90 psi (90 psi - max. available air pressure)	30 - 100 psi (100 psi - max. available air pressure)

chamfer was about 2 1/4 inches (Figure 26). As in previous tests, the hold-down straps were utilized to inhibit buckling of the specimens. The triangular geometry of the specimens, however, prevented the additional assistance of the top tension roller. After situating the electrode at a fixed height above the faying surfaces, GTA AC spot welding was initiated. When the diameter of the weld pool appeared to be approximately equal to the length of the original faying surfaces, a force on the squeeze rolls (about 40 psi) was then utilized to squeeze the coupons together.

SiC(p)/X2080 coupons having a similar geometry to the triangular-shaped 2024 coupons were prepared for GTA spot welding utilizing the MMC welding apparatus. As shown in Figure 27, a pair of 1/2 inch-wide, 90 degree-tip coupons was machined for use with each pair of triangular coupons. The 1/2 inch-wide coupons fit the 90 degree angle formed on either side of the abutting triangular coupons.

Prior to spot welding, the coupons were either grit-blasted or abraded in order to remove a lacquer that was present on the as-received composite sheet surface. The samples were degreased with acetone following the initial surface preparation. After placing a set of the four segments of a weld coupon on the copper chill block (Figure 28), GTA AC spot welding was initiated. The weld time was adjusted so the molten puddle was large enough to encompass the tips of the 1/2 inch-wide coupons, whereupon the pneumatic cylinders were actuated at a pressure level of about 40 psi. The use of the controlled geometry tipped coupons provided a means of estimating when the diameter of the weld puddle was at least equal to the original faying surface width of the triangular coupons, so that pressure could be applied at the appropriate time in order to suitably upset the joint. Additionally, the

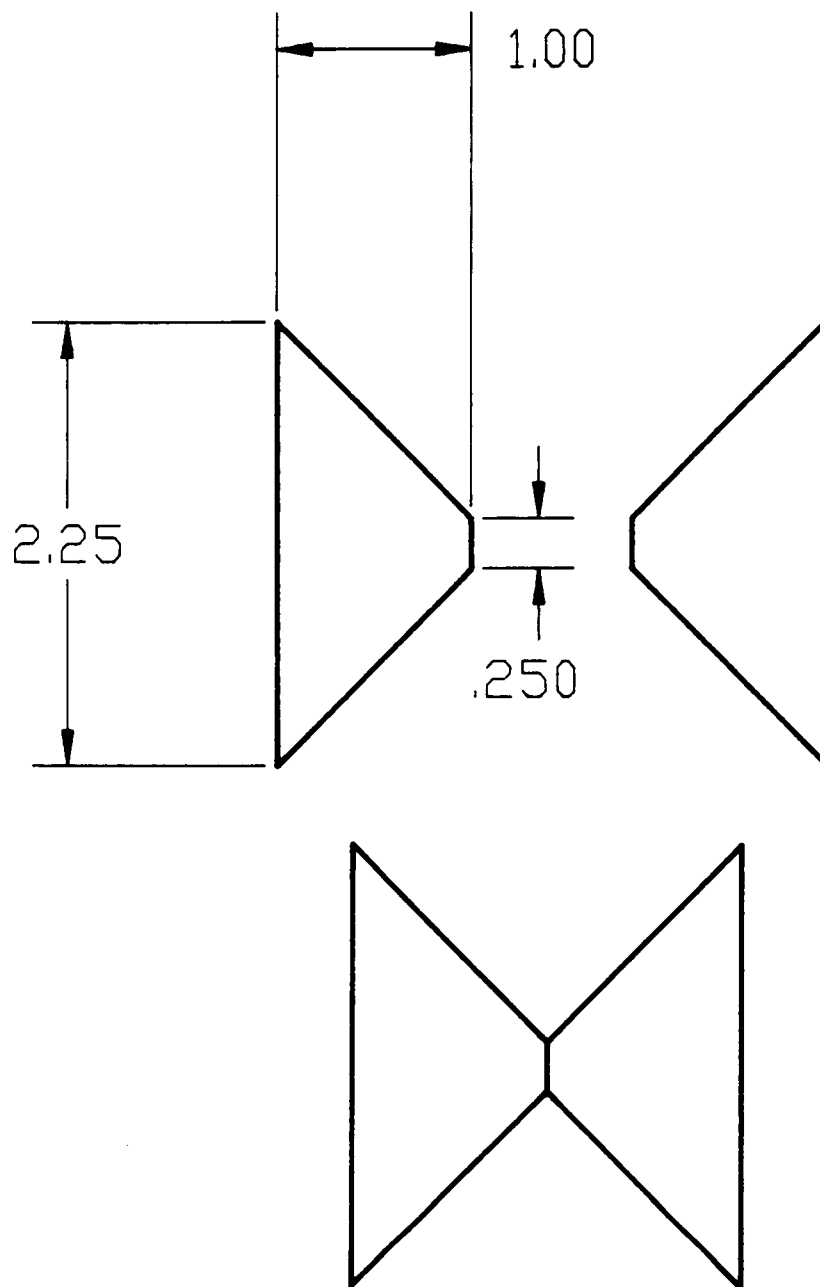


Figure 26. Schematic representation of 2024 test coupons for GTA welding in the spot mode.

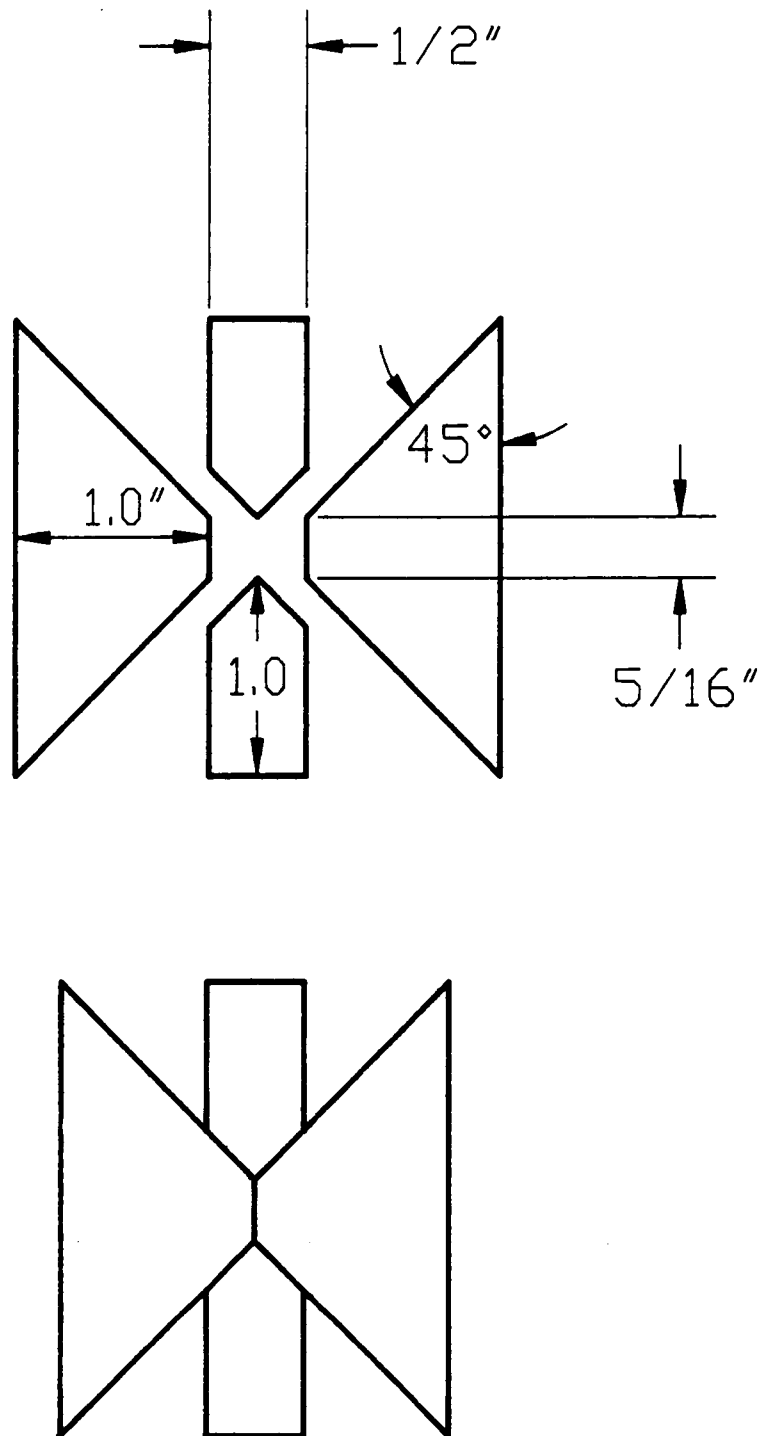


Figure 27. Schematic representation of SiC(p)/X2080 test coupons for GTA welding in the spot mode.

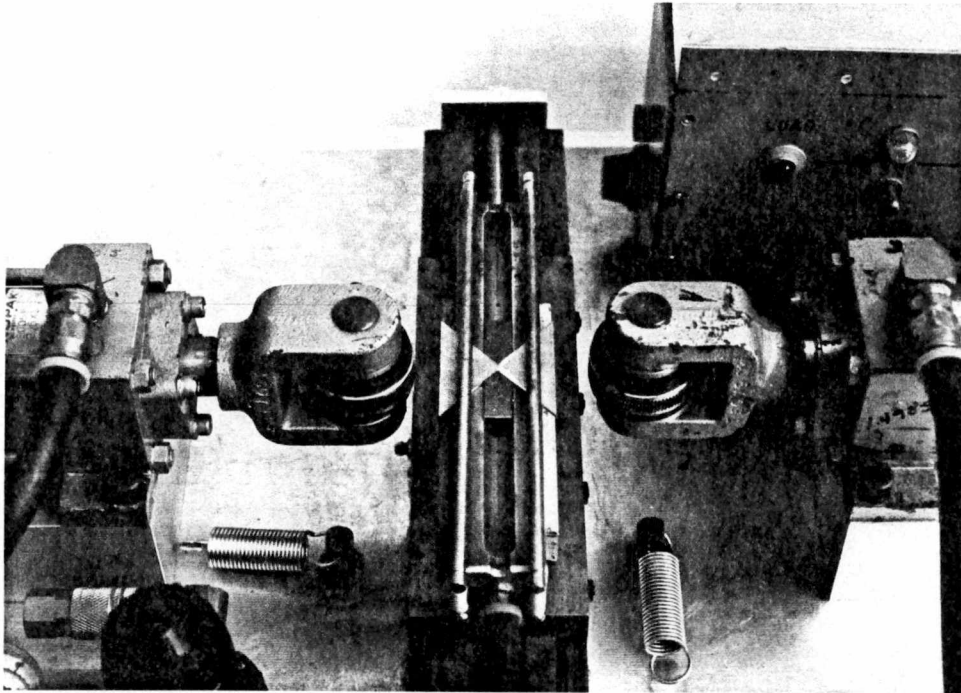


Figure 28. Set-up for GTA spot welding of SiC(p)/X2080 coupons.

tipped coupons were intended to inhibit drop-through of the molten pool that can occur due to melting at the edges of the faying surfaces.

In order to control the amount of upset due to the application of pressure, shims were placed against either side of the slide carrier. The shims were sized to reduce the amount of upset by about 1/8 inch, so that the squeeze rolls contacted the shims rather than fully advancing upon cylinder actuation. Since the total width of the triangular coupons prior to welding was 2 inches, a maximum upset reduction of about 1/4 inch could be achieved with the shims in place. The amount of upset was determined by measuring the width of the welded specimen. Based on the results of the initial tests, a second set of tests was carried out without the use of the additional tipped coupons. The parameters for GTA spot welding of the finally evolved specimens are provided in Table 9.

Following macroscopic inspection, GTA -AC spot welded coupons were selected and sectioned to obtain metallographic or mechanical test samples. Sectioning for microstructural evaluation is shown in Figure 29 (to obtain transverse profiles of the weldment). After polishing and etching the SiC(p)/X2080 samples in the same manner as the SiC(p)/2024 composites described earlier, the weldments were examined by OLM.

Specimens were also prepared for bend testing, in order to evaluate the mechanical integrity of the welds produced by using the MMC welding device. Using a 0.35 mm-thick diamond-impregnated cutting wheel, rectangular specimens containing the spot welds were sectioned from the welded coupons. The width of the bend specimens was approximately equal to the width of the original faying surfaces (5/16 inch-thick), while the length was the same as the final width of the welded coupons. For comparative purposes,

TABLE 9

**Parameters for modified GTAW of SiC(p)/X2080
in the spot mode**

Welding apparatus	Hobart arc welder
Current	90 - 100 amperes
Electrode diameter	1/8 inch
Regulated cylinder pressure	40 psi
Arc length	1/16 inch
Argon flow rate	30 CFH shielding and back purging

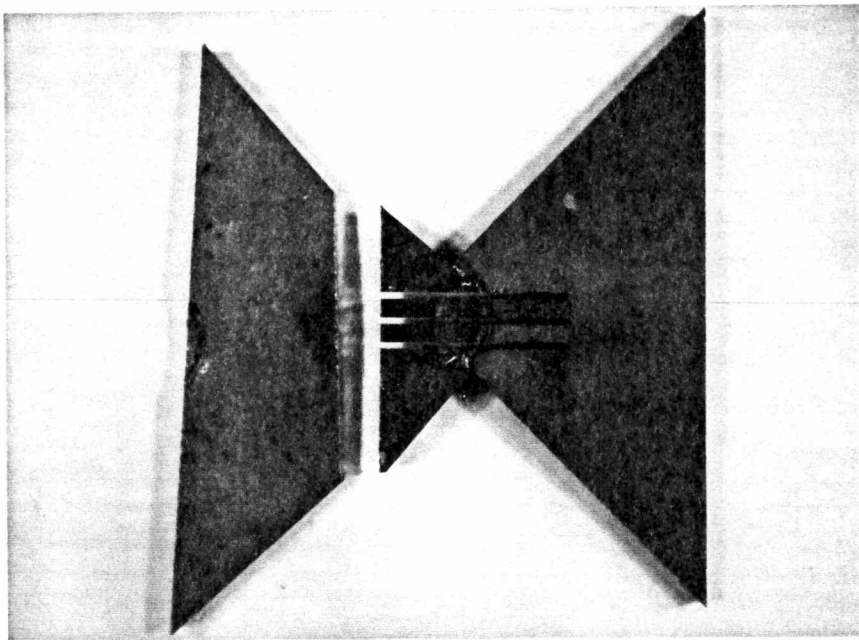


Figure 29. SiC(p)/X2080 specimen joined by modified GTAW and sectioned for metallographic inspection.

bend specimens having similar dimensions to those containing the GTA AC upset spot welds were obtained from unwelded (as-received) SiC(p)/X2080 sheet. These specimens were then either side-notched or edge-notched using the 0.35 mm-thick cutting wheel, in order to simulate defects in the spot-welded bend specimens. The remaining samples were left in the "smooth" or un-notched condition. Schematic representations of the bend specimens are shown in Figure 30.

Bend testing was conducted using a four-point flexure fixture assembly mounted on an Instron testing machine. The flexure fixture consisted of an upper bearing plate attached to the underside of a die set assembly bushing plate, and a lower bearing plate situated on a 3/8 inch-diameter ball bearing. As shown in Figure 31, the specimens were positioned between two 1 1/64 inch-diameter hardened steel loading pins on the lower bearing plate, and two loading pins on the upper bearing plate. The roller spacing on the lower plate was 30 mm, while there was a 15 mm span between the rollers on the upper plate. While a constant loading rate was utilized during subsequent testing, a load vs. elongation curve was generated in order to calculate the rupture or bending stress corresponding to the maximum load applied to the specimen.

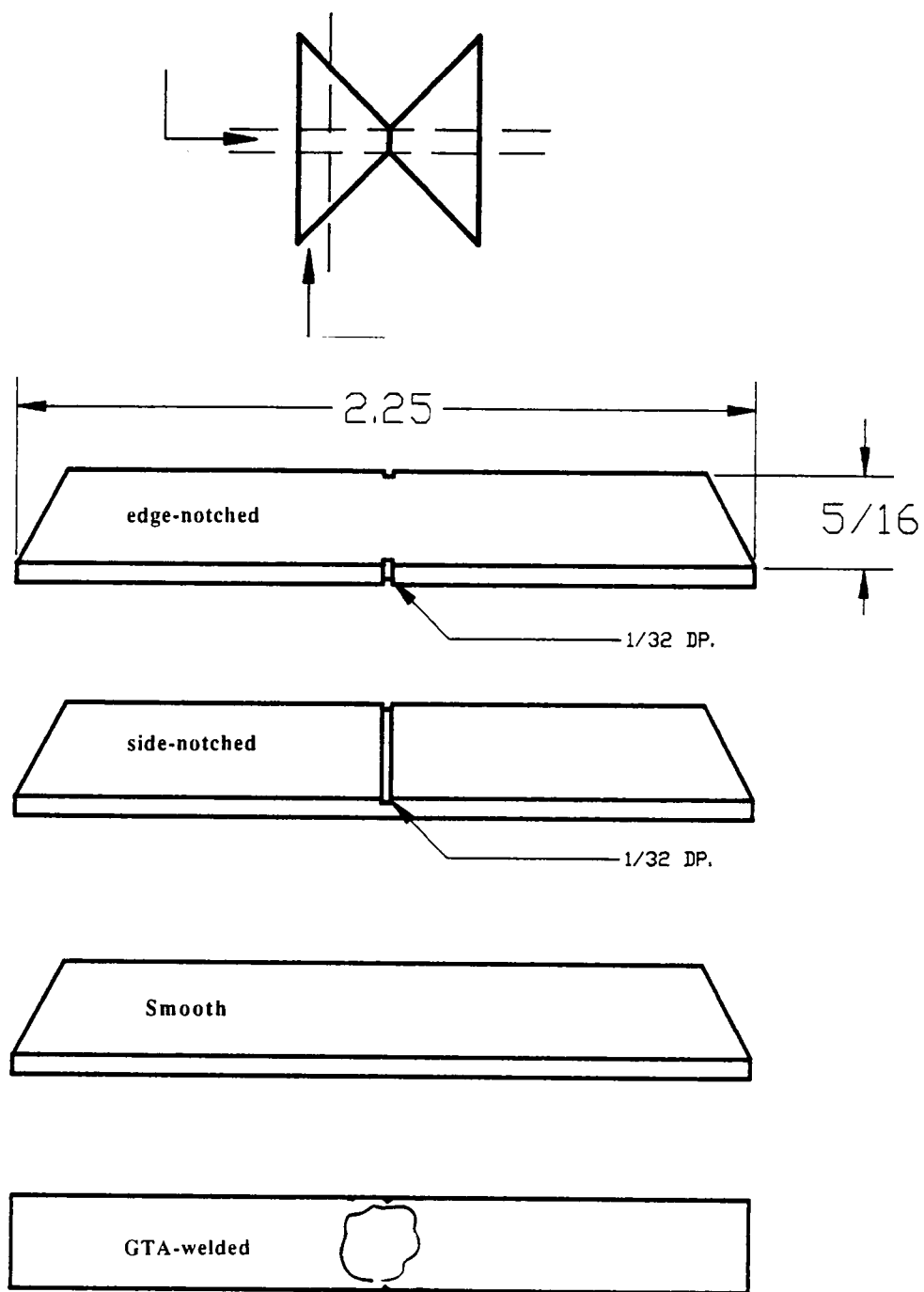


Figure 30. Schematic representation of SiC(p)/X2080 specimens subjected to four-point flexure testing.

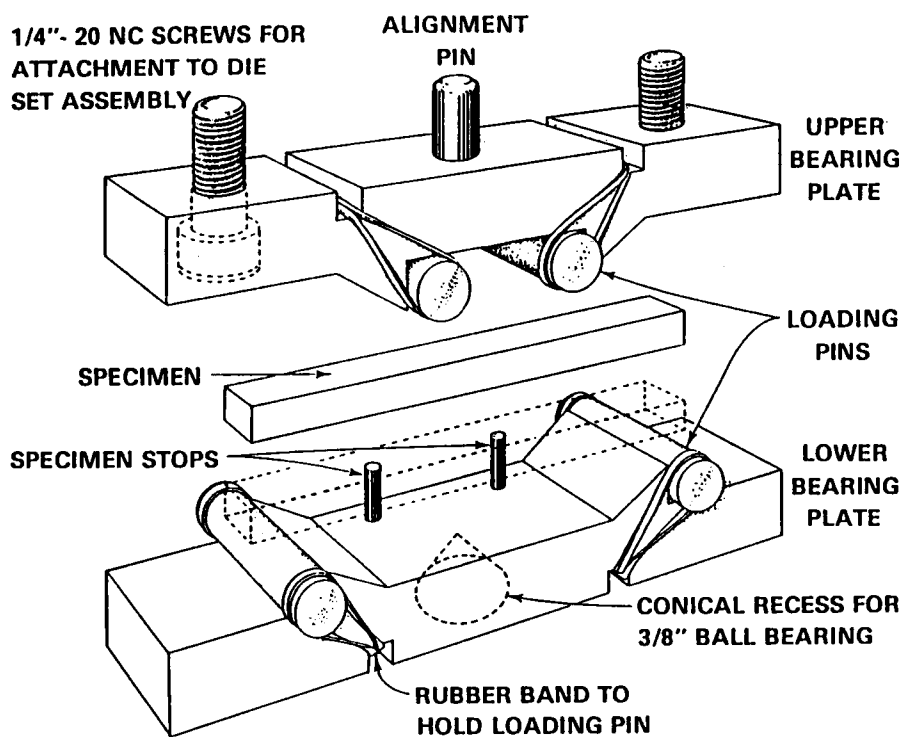


Figure 31. Four-point flexure fixture used for testing of SiC(p)/X2080.

Source: Ferber, M.K., Installation instructions, IEA annex II, Subtask 5, 4-point flexure fixture, High Temperature Materials Laboratory, Oak Ridge National Laboratory, March 1990.

CHAPTER IV

RESULTS AND DISCUSSION

Evaluation of LB welds in SiC(p)/2024

As described in chapter three, beads produced by LB processing of 15 vol % SiC(p)/2024 at various heat input levels were sectioned and examined macroscopically and by optical light microscopy (OLM). The quality of the welds appear to be acceptable based on initial macroscopic examination, as adequate penetration without gross porosity is achieved for all heat input levels and shielding methods. As predicted (based upon observations in a previous study involving LBW of Al alloys), the depth of penetration is slightly improved when only coaxial gas is used as opposed to both coaxial and supplemental shielding gas [60].

The typical unetched microstructure of the LB welds is similar to that shown in Figure 32. While the majority of the fusion zone and the base material exhibit similar unetched appearances based on the distribution of SiC particulates, the characteristic LB “keyhole” contains a markedly different microstructure. A fine dispersion of an acicular constituent is present in this region, in addition to a few “surviving” SiC particulates in some of the welds (Figure 33). Further examination of the microstructural profile reveals a relatively limited area adjacent to the keyhole in which partial dissolution of SiC is evident. As shown in Figure 34, the silicon carbide particulates at the keyhole boundary appear to serve as nucleation sites for the acicular particles.

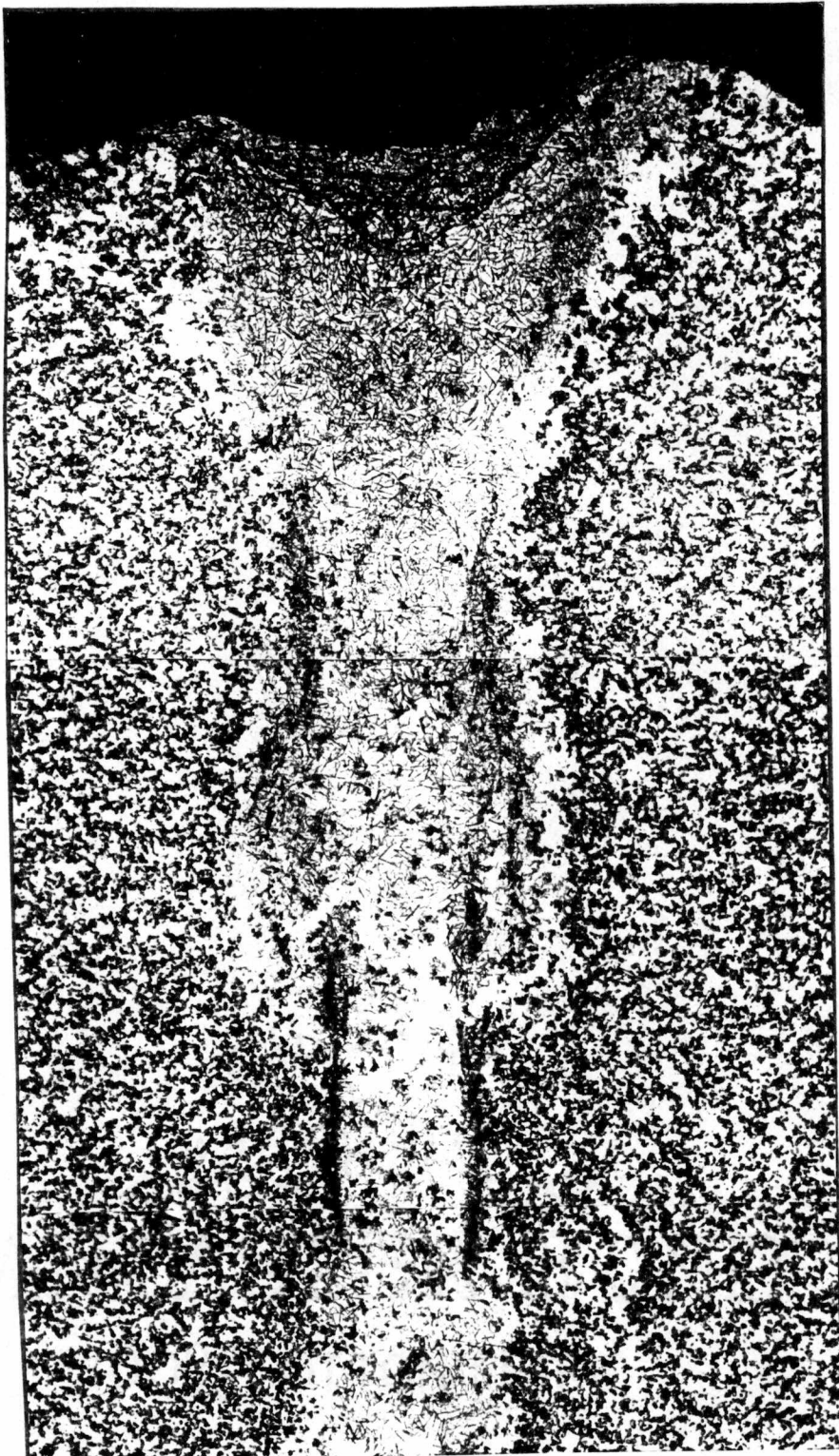


Figure 32. Typical microstructural appearance of LB-welded SiC(p)/2024 (unetched, 50X).

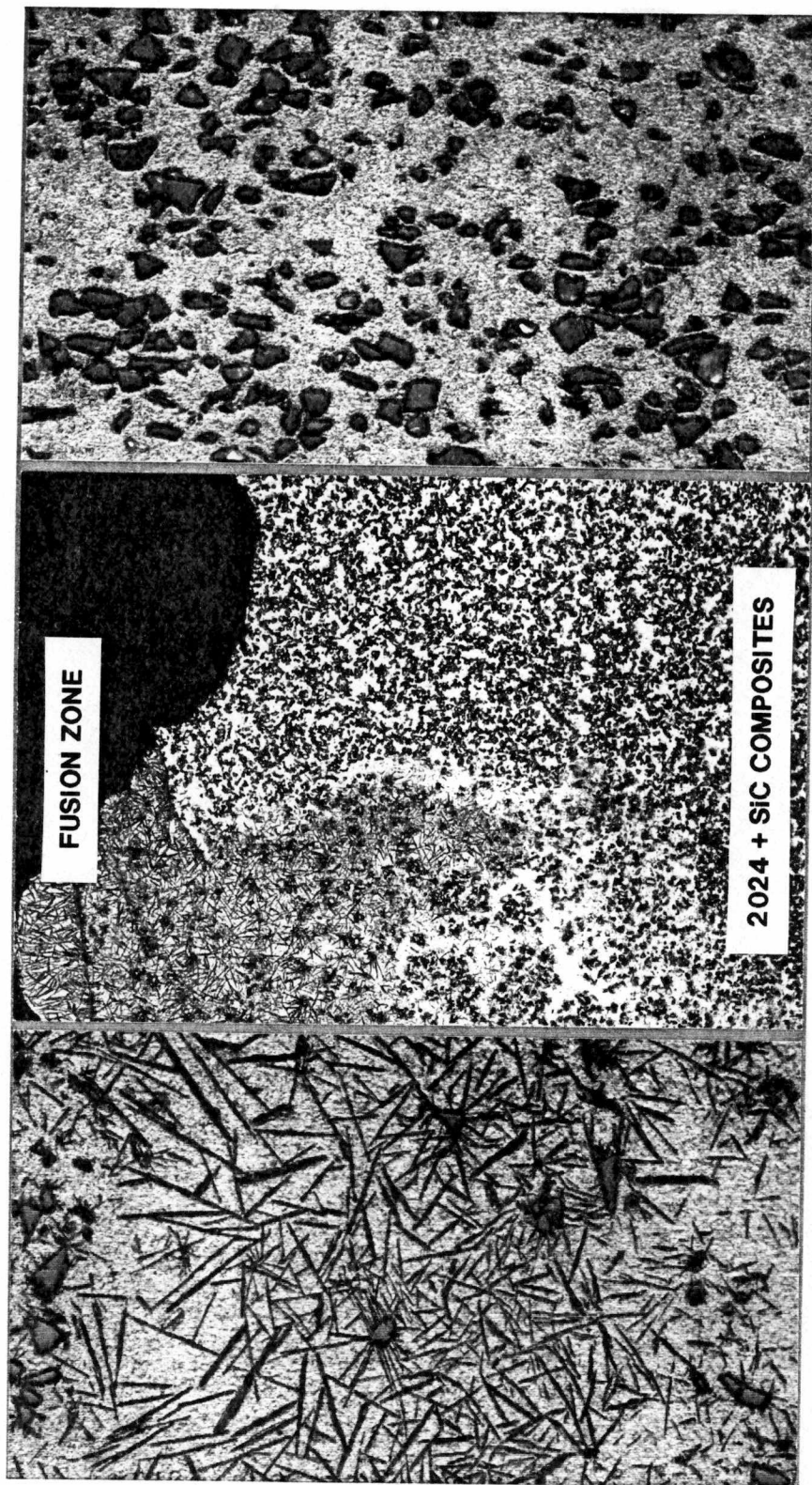


Figure 33. Comparison of keyhole microstructure produced by LBW of SiC(p)/2024 (left, 400X) to base material microstructure (right, 400X). The fusion zone at low magnification is shown at center (50X).

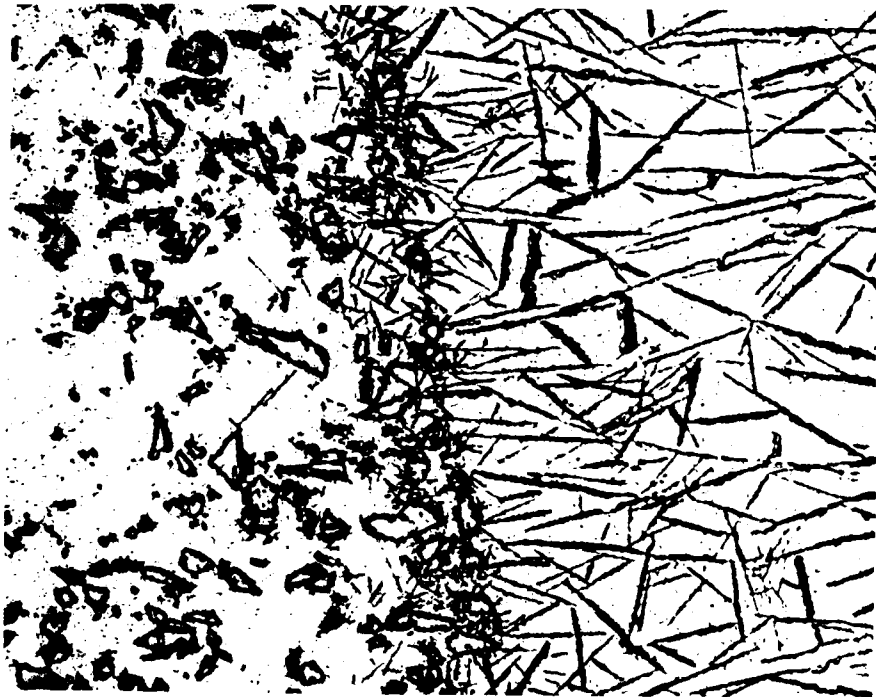


Figure 34. Boundary between keyhole and "outer" fusion zone in LB-welded SiC(p)/2024 (400X).

Examination of the etched LBW microstructures confirms that the fusion zone consists of at least three distinct regions. While a similar underlying solidification structure is present in each of these regions, the distribution of the reinforcing particulates varies throughout the fusion zone. The keyhole is characterized by an absence of SiC particulates and the presence of an acicular or needle-like constituent, where the size and density of this constituent appears to decrease with distance from the center of the keyhole. Based on the results of EDS analysis (Figure 35), the acicular phase is shown to be aluminum-rich (carbon could not be analysed). Adjacent to the keyhole, an “intermediate” zone exhibits partial SiC dissolution, while an “outer” fusion zone retains a base material SiC distribution. The microstructures at various locations relative to the center of the LB fusion zone in SiC(p)/2024 are shown in Figures 36 and 37, and the appearance of the keyhole as revealed by SEM is shown in Figure 38.

The results of microhardness testing show that peak hardness levels of as-welded specimens are attained at the center of the keyhole, while hardness steadily decreases throughout the fusion zone to a constant value in the base material. As shown in Figure 39, keyhole hardness levels are higher than 200 DPH, whereas base material levels do not exceed about 85 DPH. Intermediate hardness values are exhibited in the outer fusion zone. Normally, some softening would be expected to occur in the heat affected zone (HAZ) of dispersion-hardenable Al alloys, due to dissolution of strengthening precipitates in this region. The fact that the microhardness profiles reveal no significant evidence of this effect can be attributed in part to the minimal HAZ produced by LBW. HAZ softening is also counteracted by internal stresses generated due to thermal expansion differences between the SiC reinforcements and the matrix alloy - a factor which contributes to the increased hardness in the outer fusion zone, as well.

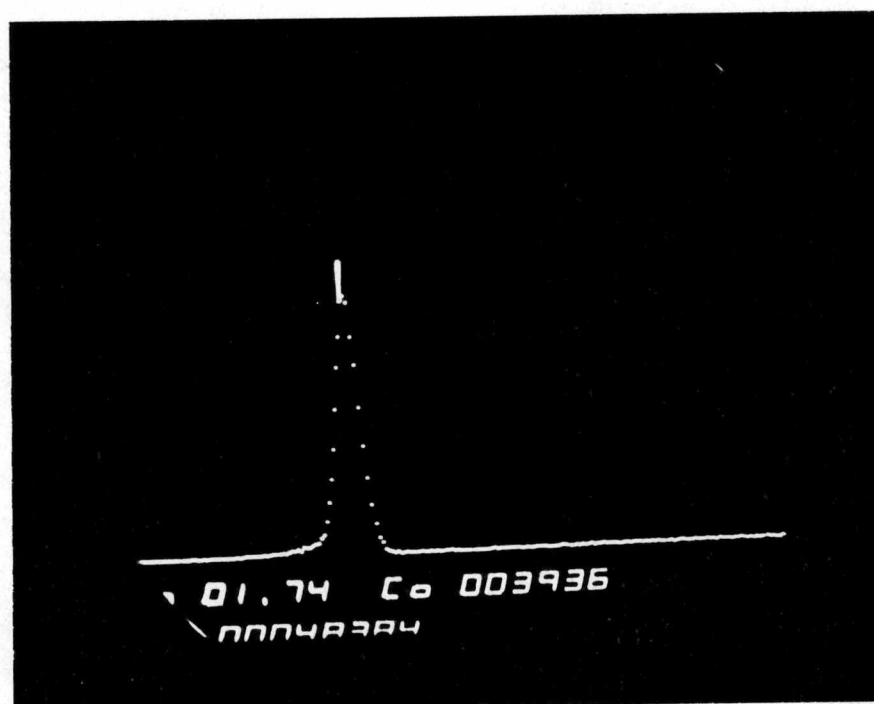
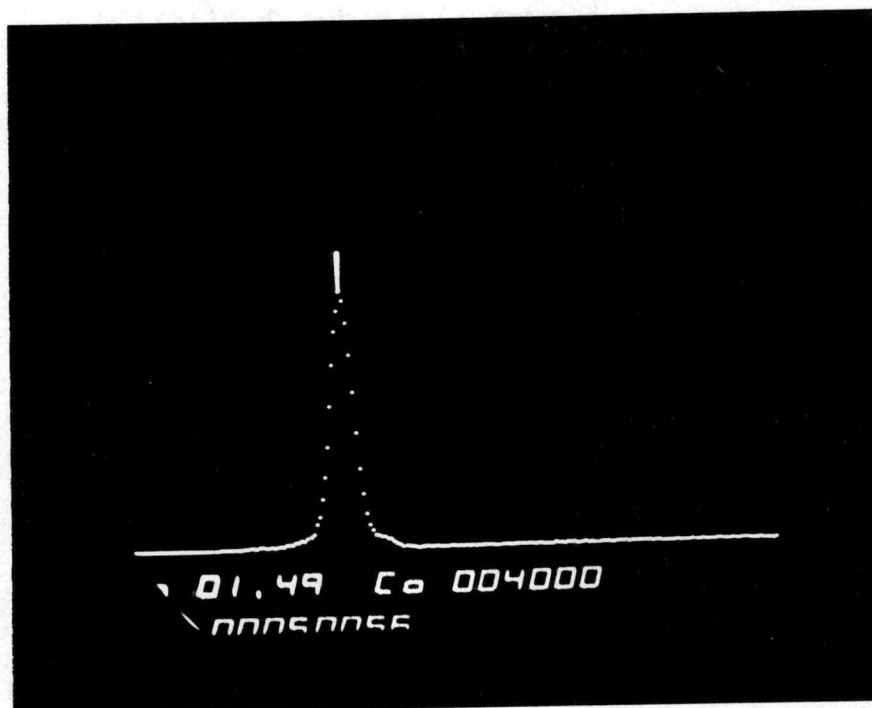


Figure 35. Characteristic Al (top) and Si (bottom) peaks produced by EDS analysis of acicular particles and SiC particulate, respectively.

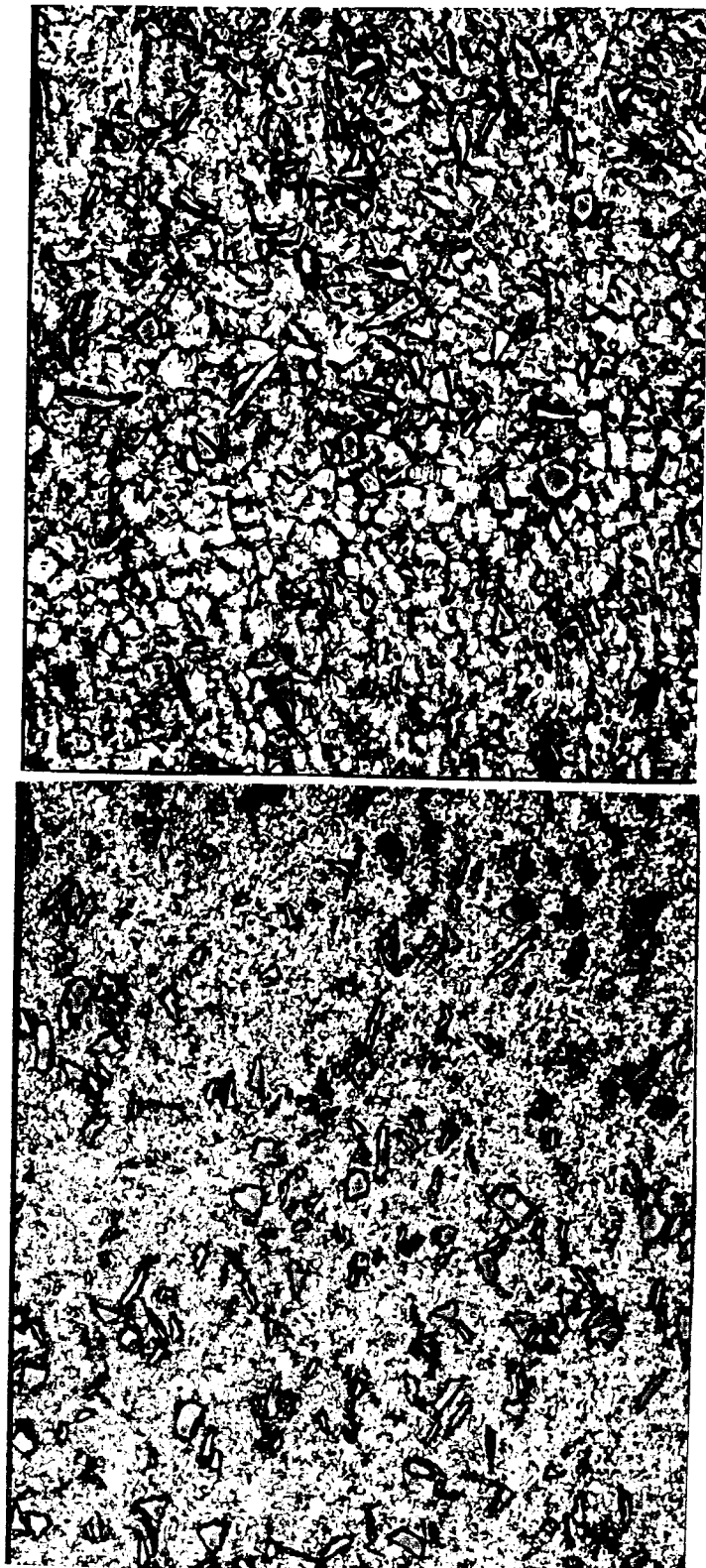


Figure 36 Base material (left) and "outer" fusion zone microstructure (right) in SiC(p)/2024 (400X).

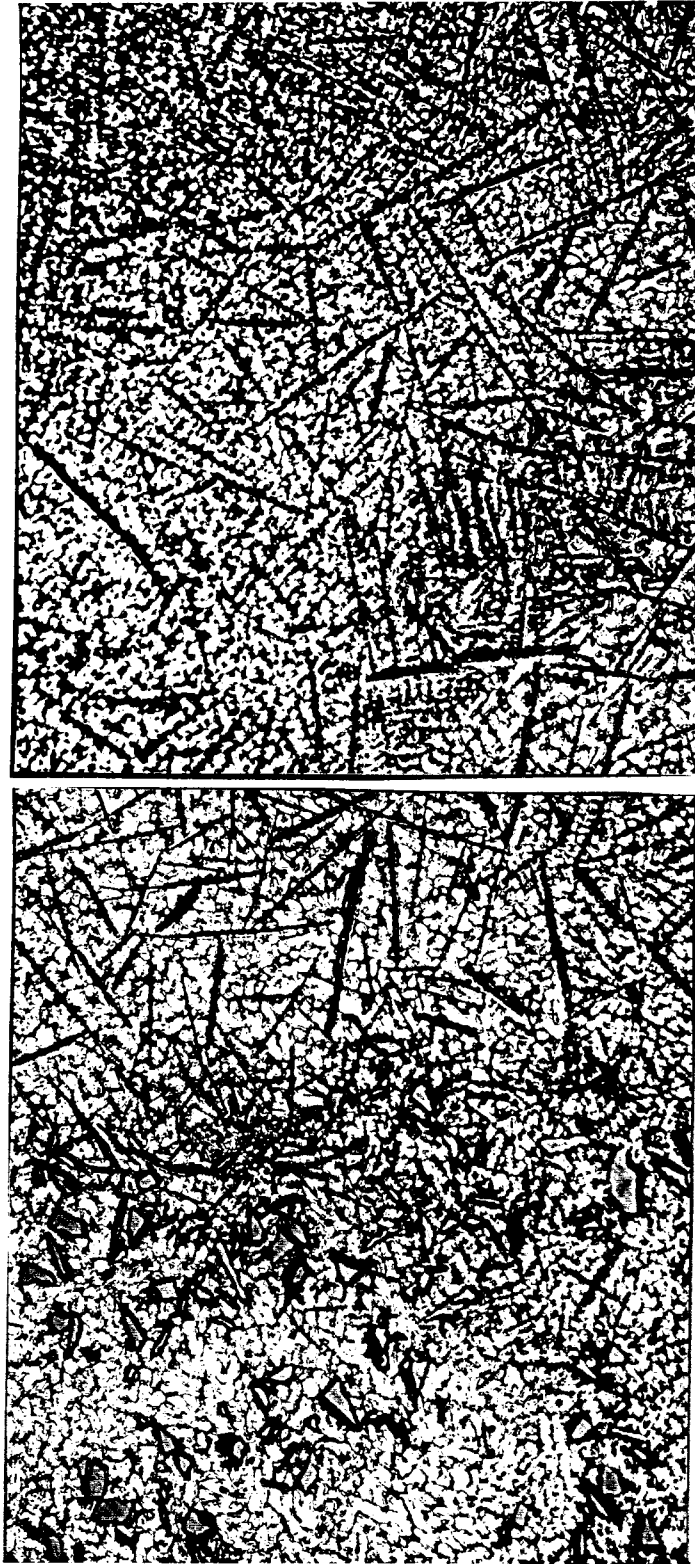


Figure 37. Intermediate fusion zone (left) and keyhole microstructure (right) produced by LBW of SiC(p)/2024 (400X).



Figure 38. Keyhole region in LB welded SiC(p)/2024 (1000X).

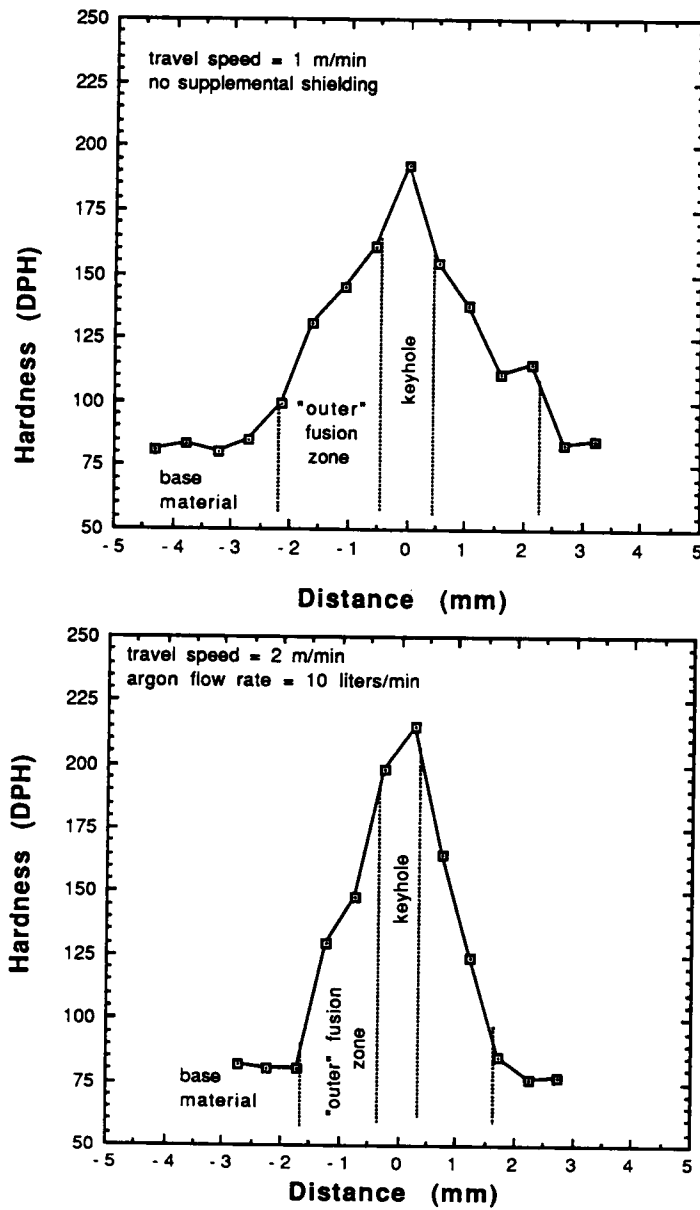


Figure 39. Variation in hardness with distance from center of LBW in SiC(p)/2024.

Postweld heat treating at 300 °F for 24 and 72 hours has little effect on the hardness of either the base material, outer fusion zone, or keyhole, as shown in Figure 40. After postweld heat treating several specimens at 400 °F, reductions in the fusion zone and base material hardness are observed, which are most likely the result of matrix overaging. The general shape of the microhardness profile is retained, however, with the keyhole being significantly harder than the outer fusion zone and base material. Metallographic examination of the heat-treated samples by OLM revealed no obvious microstructural changes.

Tensile testing of dogbone specimens containing transverse sections of LB welds in SiC(p)/2024 led to fracture at the center of the fusion zone in all specimens. As shown in Figure 41a, examination of tested specimens reveals no obvious cross-sectional area reduction or gauge length elongation; rather, specimen fracture surfaces show evidence of brittle, cleavage-type fracture (Figure 41b). Typical tensile strength values given in Table 10 for samples containing transverse and “top” (parallel to the plate surface) sections of LB beads are 253 MPa (36 ksi) and 358 MPa (51 ksi), respectively.

Immersing LB welded specimens in pure water revealed that the weldment is highly susceptible to aqueous attack - a characteristic which can be attributed to the presence of water-soluble aluminum carbide. After exposure of a polished sample to water for several days, the fusion zone suffered severe deterioration. The keyhole appeared to have experienced preferential chemical attack, exhibiting an irregular surface and no recognizable microstructural features.

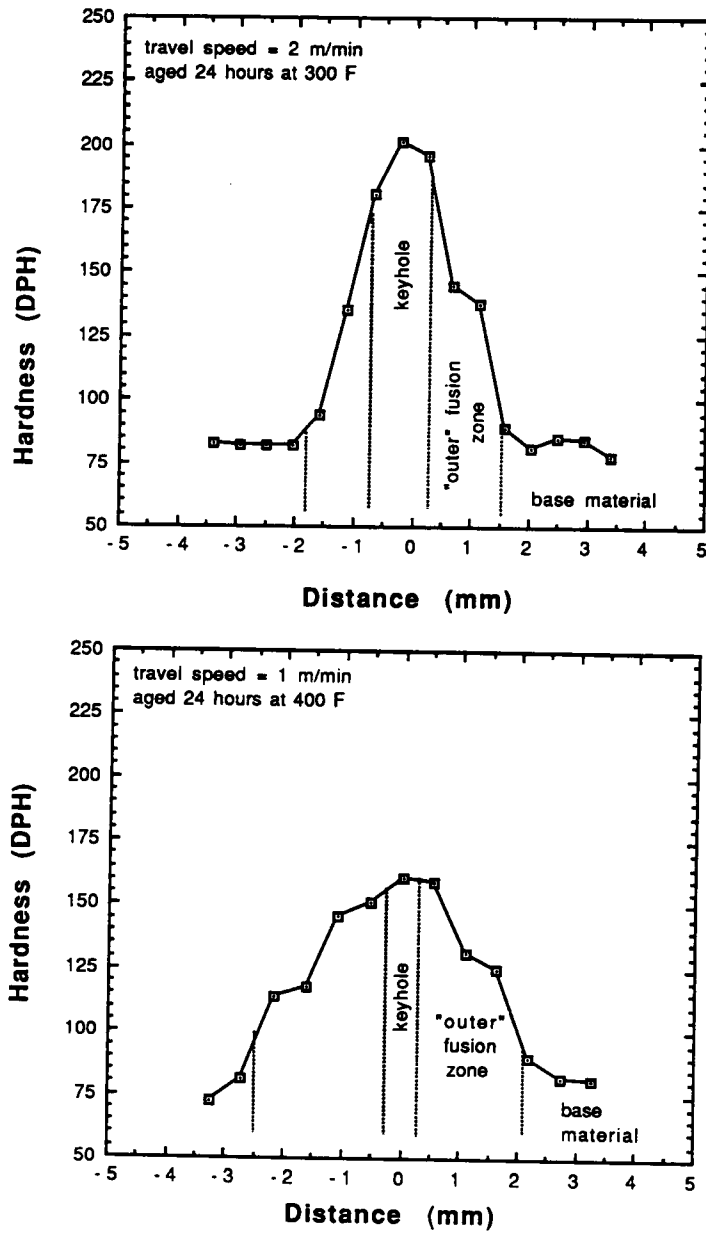


Figure 40. Variation in hardness with distance from center of LBW in SiC(p)/2024, following aging at 300 °F (top) and 400 °F (bottom).

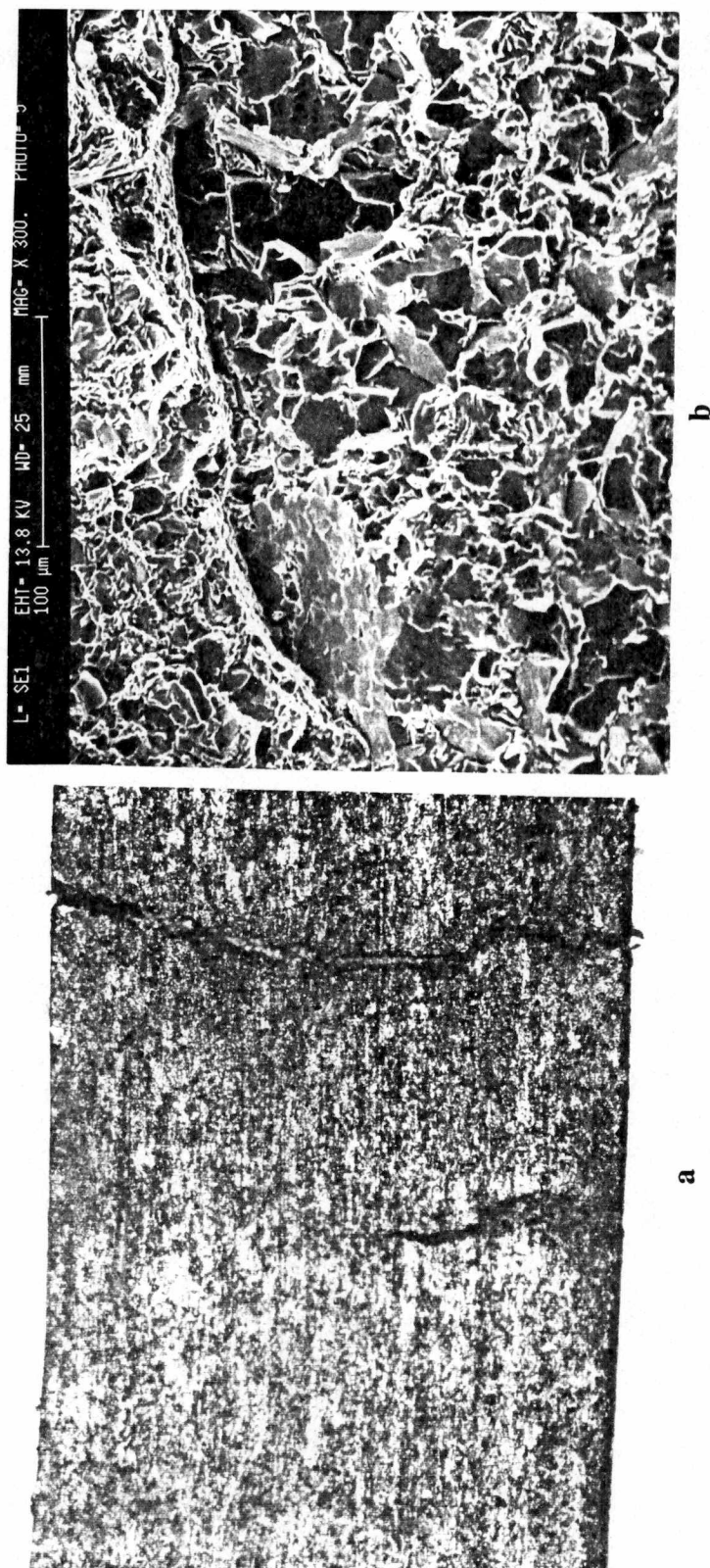


Figure 41. Fractured SiC(p)/2024 "dogbone" tensile specimen (left) as shown by OLM (left, 50X) and corresponding fracture surface as examined by SEM (right, 300X).

TABLE 10**Typical tensile properties of LB-welded SiC(p)-reinforced 2024**

Specimen (as shown in Figure 21)	“A” type	“B” type
Cross-sectional area along gauge length	$2 \times 10^{-3} \text{ in}^2$ (0.013 cm ²)	$2 \times 10^{-3} \text{ in}^2$ (0.013 cm ²)
Crosshead speed of Instron tensile testing apparatus	0.01 in/min (0.025 cm/min)	0.01 in/min (0.025 cm/min)
Ultimate tensile strength	51 ksi (358 MPa)	36 ksi (253 MPa)

Based on the agreement of the results just described with those of the studies discussed in chapter two, the needle-like particles observed in the keyhole of LB welded SiC(p)/2024 are determined to be aluminum carbide (Al_4C_3). The properties observed for LB welds in this study, for example, are consistent with those noted in previous studies on SiC-reinforced Al alloys; that is, the presence of the needle-like or plate-like Al_4C_3 imparted high hardness and brittle behavior, and high corrosion susceptibility in aqueous environments. Additionally, the formation of the carbide corresponds to the loss of SiC and hence the engineered composite properties. The inability to improve the weld properties by post weld heat treating further reinforces the determination that the plate-like constituents in the keyhole are aluminum carbide, since the Al-C system contains only a single intermediate compound, Al_4C_3 , which decomposes only at temperatures in excess of 2156 °C (Figure 42, [38]).

LBW of “Nicalon”(f)/1050

LBW of Nicalon fiber-reinforced aluminum was carried out by both continuous CO_2 and pulsed Nd:YAG processing, as described in the previous chapter. Relative to LBW of SiC(p)/2024, lower heat inputs were utilized for welding the continuously reinforced composites. However, N_2 shielding gas was used rather than argon with the objective of improving energy absorption by promoting the formation of aluminum nitride on the molten pool during welding [62]. Both pulsed and continuous LBW, however, proved to be ineffective with respect to this composite at any combination of travel speed and current. According to XDE Systems, Inc., a plume could not be established during welding for any adjustment of the laser focusing elements or variation in travel speed. The difficulty during bead-on-plate welding is reflected in the poor quality of the resultant

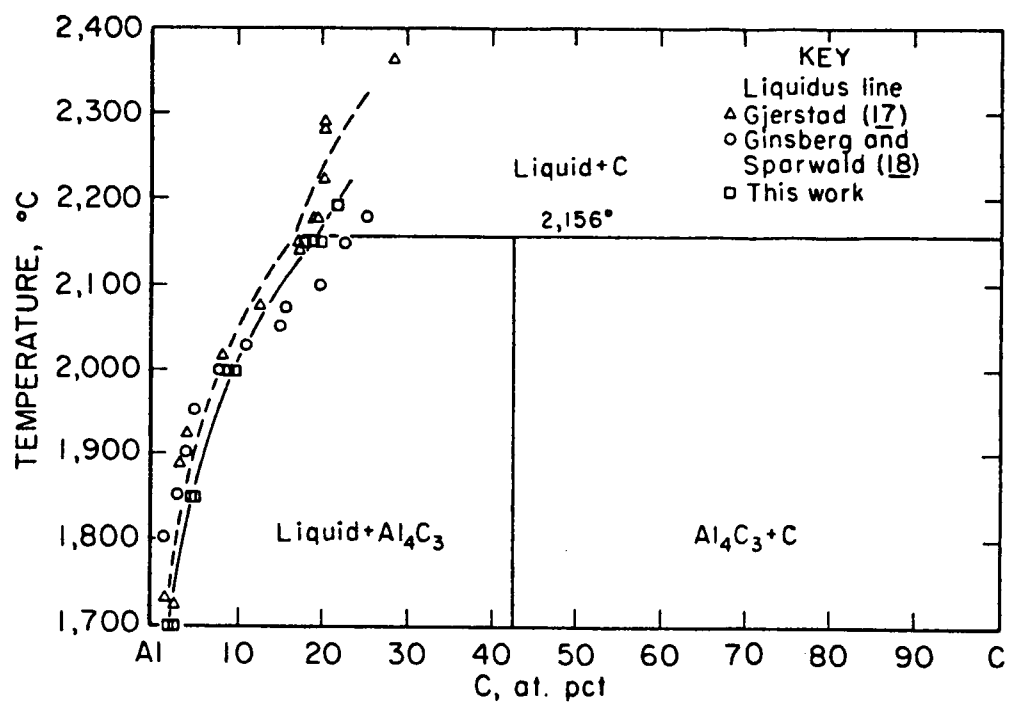


Figure 42. The binary system Al-C.

Source: Oden, L.L, and McCune, R.A., Phase equilibria in the Al-Si-C system, *Metallurgical Transactions* 18 A, December 1987, 2005-2014.

welds, which exhibit gross porosity based upon inspection with the unaided eye. Subsequent metallographic inspection by OLM reveals the presence of Al_4C_3 particles, whose size and density increase with increasing heat input. Although the quality of all welds produced in these composites is unacceptable, somewhat better results are achieved when Nd:YAG welding is conducted in the direction parallel to the reinforcing fibers, rather than CO_2 welding in the transverse direction. Aluminum carbide formation, for example, is reduced in the Nd:YAG welds. Photomicrographs corresponding to OLM evaluation of Nicalon fiber-reinforced aluminum are shown in Figure 43 and 44.

Results of GTA welding of SiC(p)/2024, SiC(p)/X2080, and graphite/6061

GTAW carried out at various current levels also produces welds in which a portion of the fusion zone is characterized by the dissolution of SiC particulates and formation of aluminum carbide, while the outer fusion zone retains the appearance of the unetched base material. As shown in Figure 45, higher heat input corresponds to an increase in the size and density of Al_4C_3 plates in SiC(p)/2024 standing edge welds. This effect is even more apparent when the Al_4C_3 particles produced by GTAW are compared to those formed due to the lower overall energy input of LBW (Figure 32). In addition to the more extensive formation of aluminum carbide, porosity is also higher in GTA welds than in LB welds. The etched microstructural profile in standing edge-welded SiC(p)/X2080 is shown in Figure 46, where the "upper" portion of fusion zone associated with higher temperatures consists of Al_4C_3 needles, blocky silicon, and the underlying cellular eutectic structure. As in the LB welds, an "outer" fusion zone is also present in SiC(p)/X2080, which exhibits a cellular eutectic structure and a distribution of SiC similar to that of the base material due to the lower temperatures encountered relative to those in the "upper" fusion zone .

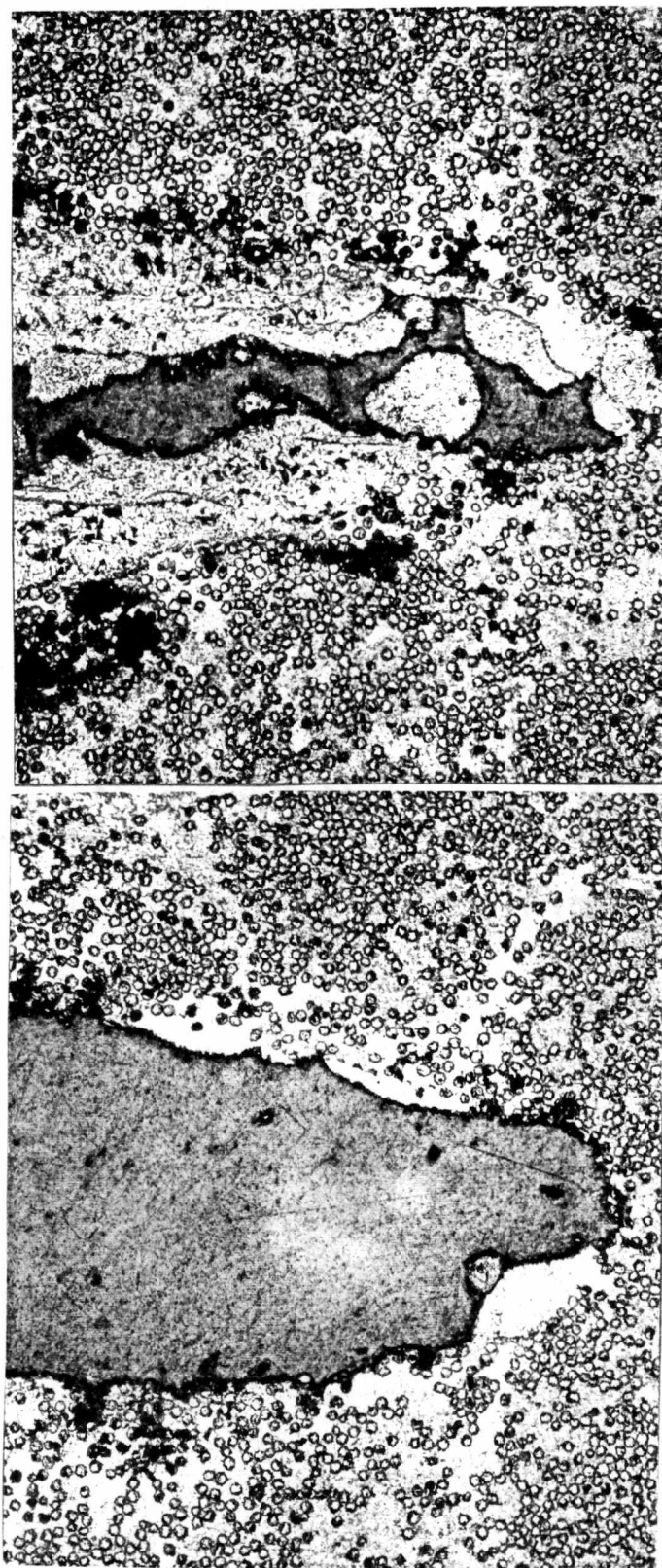


Figure 43. Cavities produced by Nd:YAG LB welding of SiC(f)/1050 in the "longitudinal" direction (100X).

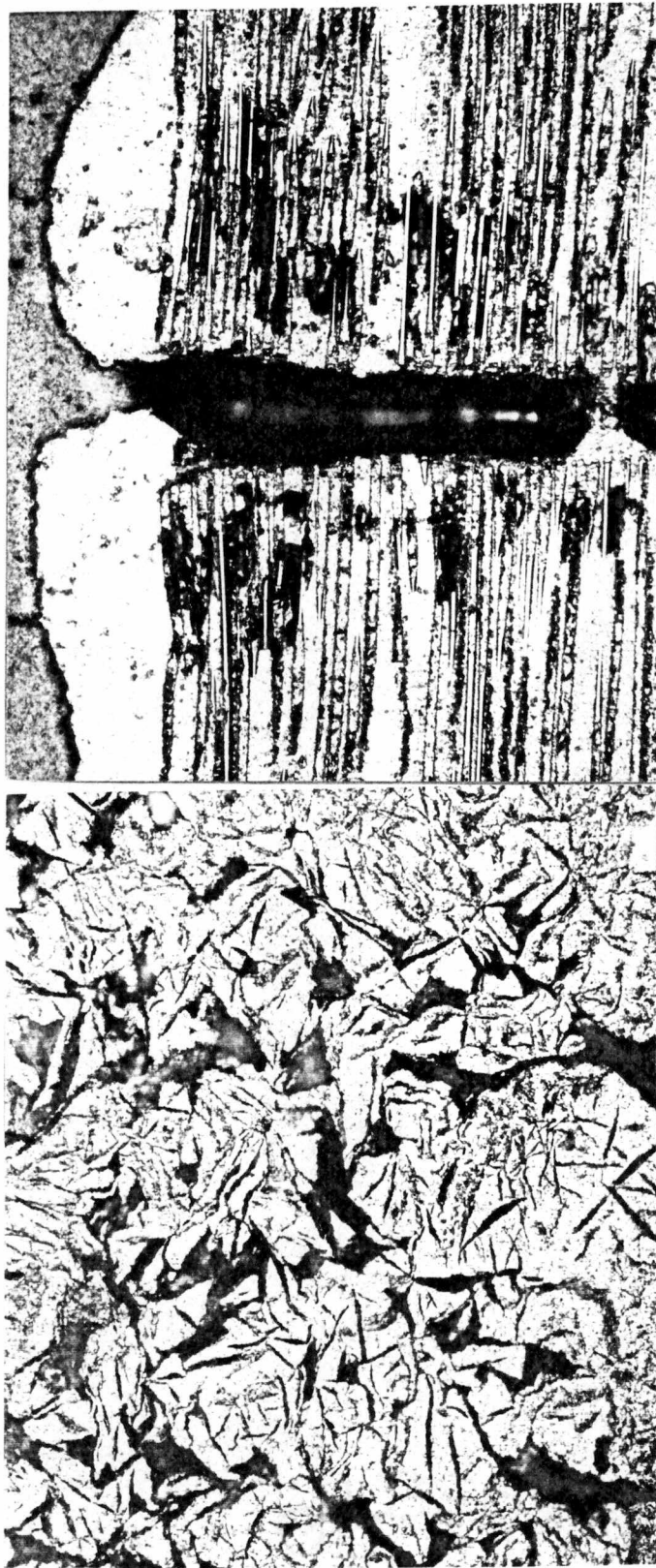


Figure 44. Cavity produced by CO₂ LB welding of SiC(f)/1050 in the transverse direction (right, 100X), and Al₄C₃ particles in material adjacent to the cavity (left, 400X).

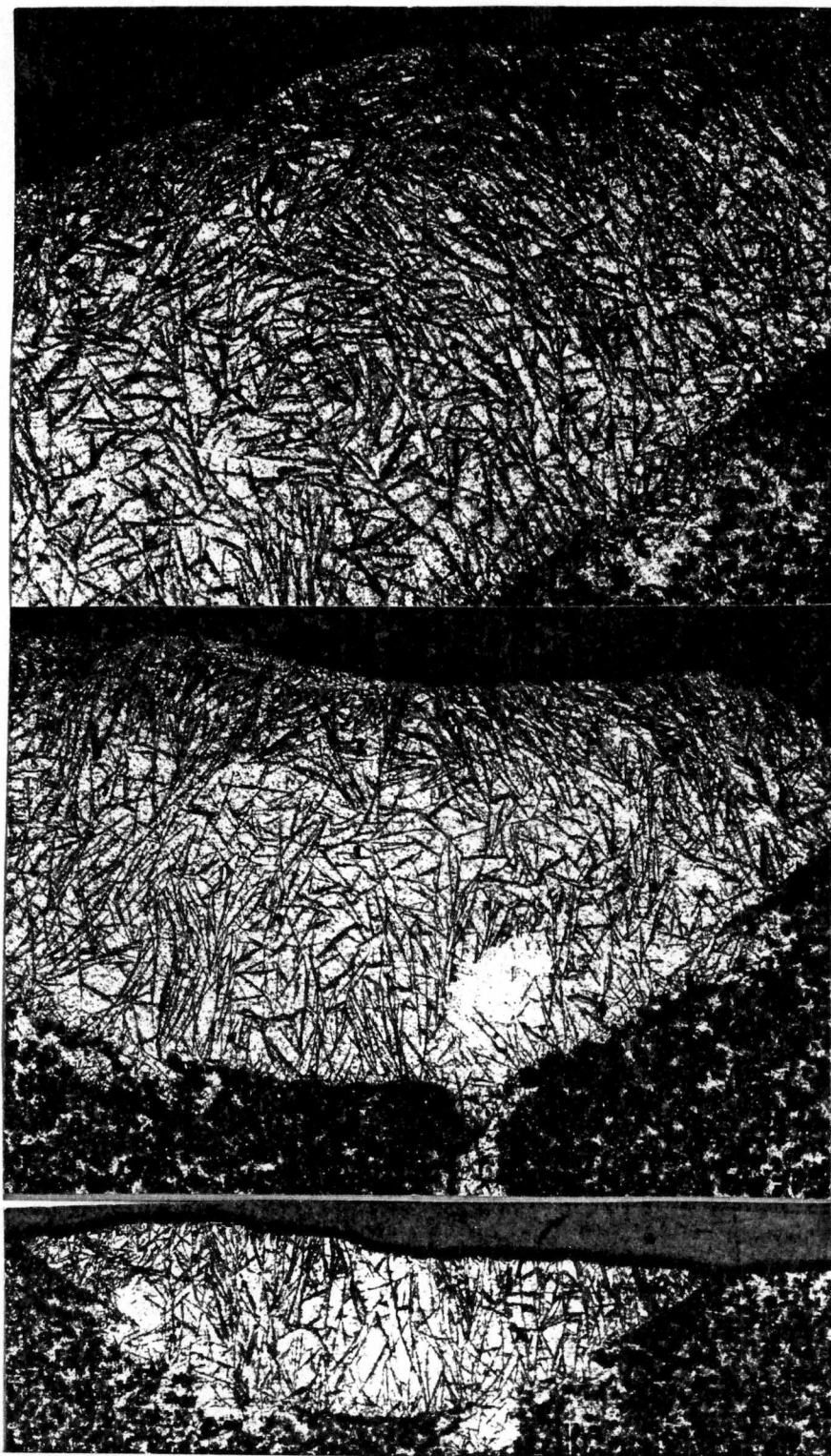


Figure 45. Transverse profiles of standing edge welds produced by GTAW of SiC(p)/2024 at 50X, corresponding to current levels of approximately 115 (top), 140 (center), and 200 amperes (bottom).

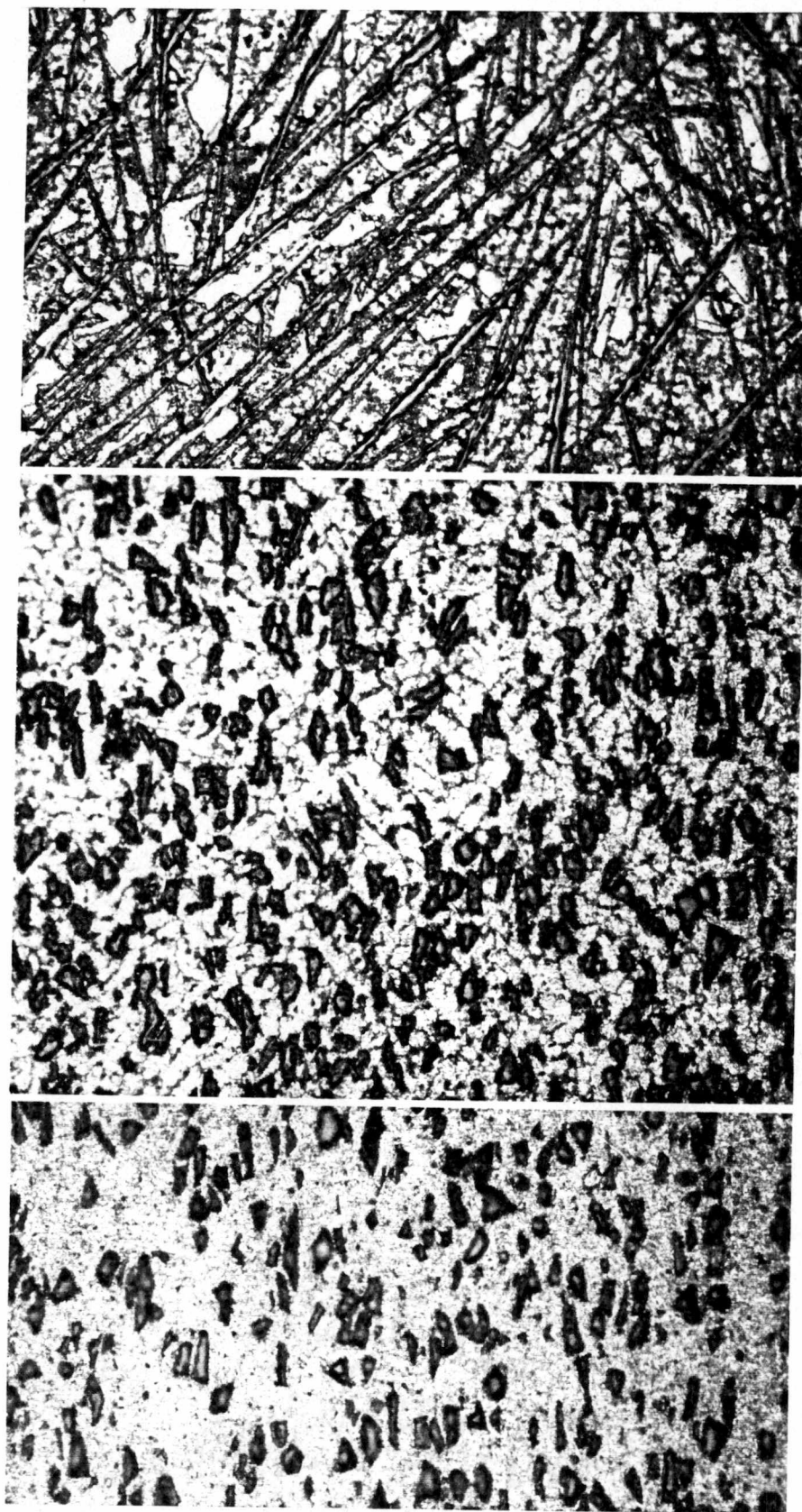


Figure 46. Base material (left), "outer" fusion zone (center), and "upper" portion of fusion zone (right) in SiC(p)/X2080 specimen subjected to GTA standing edge welding (400X).

SEM examination of fracture surfaces obtained from the standing edge welds reveal evidence of two distinctly different fusion zones associated with GTA welding of the SiC/2024. Since the depth of weld penetration was minimal relative to the specimen height, fracture through the fusion zone was easily induced by inserting a wedge between the plates and then prying them apart to obtain a longitudinal profile of the weldment. A transverse profile was subsequently obtained by placing a deep sawcut in one of the separated plates and striking the plate until a crack initiated at the base of the sawcut and propagated through the fusion zone. As shown in Figure 47, low magnification examination of the fracture surface in the transverse direction reveals two distinct fracture patterns within well-defined boundaries. Both of these patterns are contained within regions having "weld-shaped" configurations, where one region is associated with higher energy input during welding. In the "upper" portion of the fusion zone, the fracture surface is characterized by brittle, cleavage-type fracture. This fracture pattern is similar to that observed earlier for the tensile specimens, in which failure occurred in the welds produced by LBW. Thus, the mode of failure can be related to the presence of the brittle Al_4C_3 .

Adjacent to the upper portion of the fusion zone, the surface features change dramatically from flat, cleaved planes to a dimpled structure characteristic of ductile failure. A significant amount of porosity is also present in this area, as seen in Figure 48. Higher magnification examination of the pore surfaces shows evidence of melting and resolidification, which indicates that this region is an "outer" fusion zone. Provided that fusion occurs in this area, the ductile-type failure mode implies that formation of Al_4C_3 is inhibited while SiC particulates remain intact. The remainder of the fracture surface

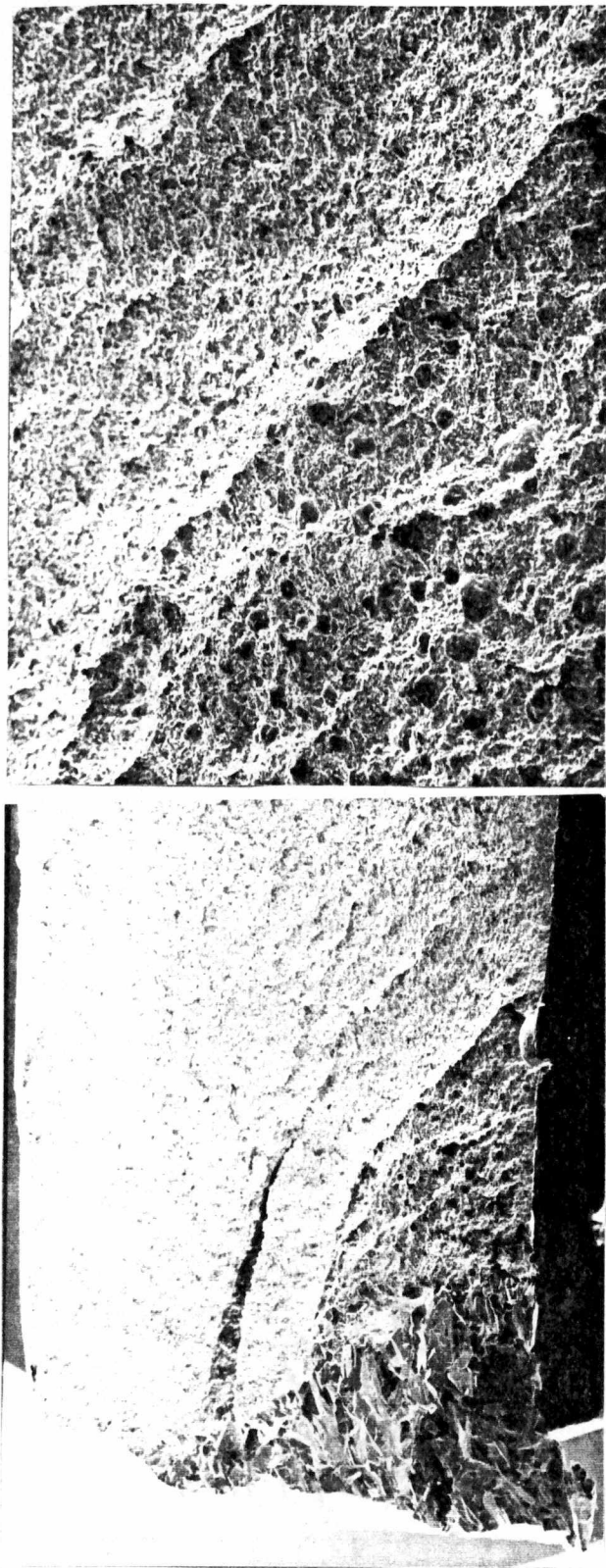


Figure 47. Fracture specimen obtained from GTA-welded SiC(p)/2024, shown at 20X (left) and 60X (right).

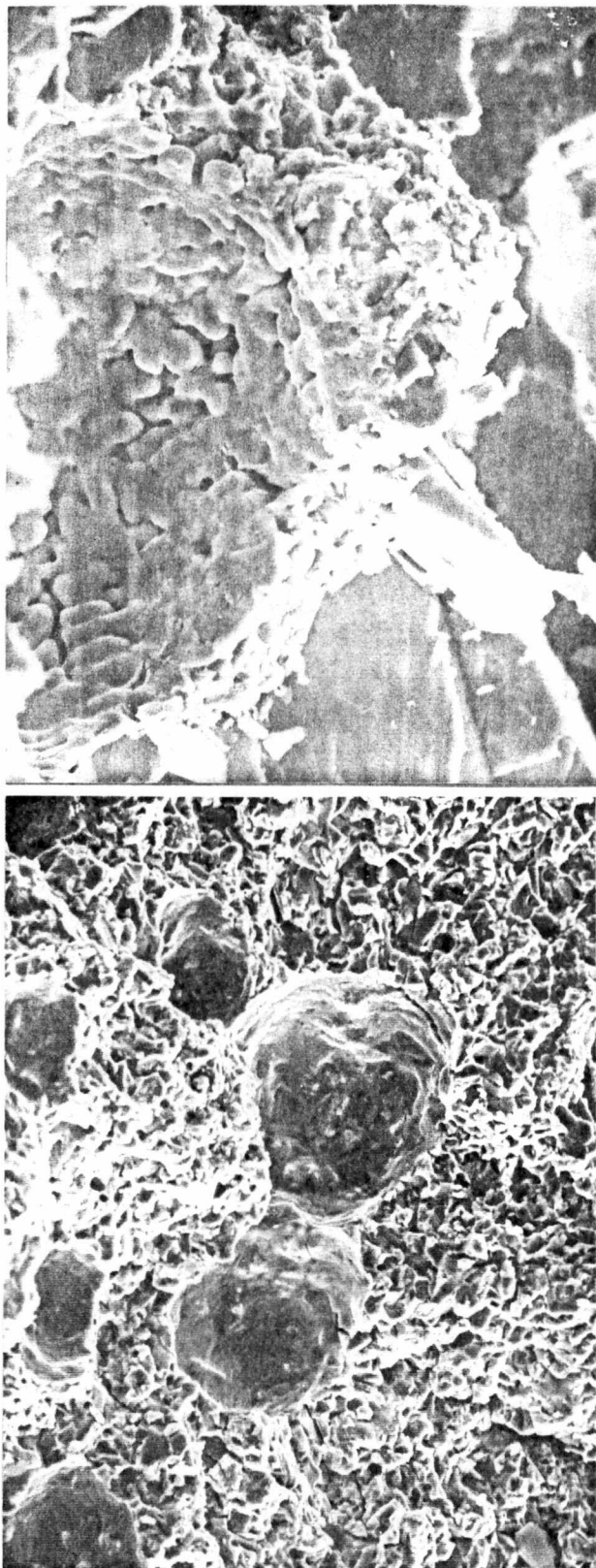


Figure 48. Fracture pattern associated with "outer" fusion zone in GTA-welded SiC(p)/2024, shown at 300X (left) and 660X (right).

(corresponding to the base material) exhibits the same ductile-type fracture pattern as the “outer” fusion zone, although there is no porosity evident. Excluding the effect of porosity, the similar fracture patterns in the “outer” fusion zone and base material imply a corresponding similarity in mechanical properties. Thus, in the absence of the “upper” portion of the fusion zone in which aluminum carbide formation occurs, relatively uniform properties from the base material through the weldment can be expected.

Metallographic inspection of the graphite-reinforced 6061 specimen subjected to spot welding reveals the presence of large, acicular particles that are most likely Al_4C_3 . After exposure to air for several days, a powdery corrosion product formed in the weld region, after which disintegration of the fusion zone occurred. Surrounding material experienced massive degradation as well, due to the separation of reinforcing fibers from the matrix material. Thus, the severe corrosion susceptibility and microstructural appearance of the spot weld in graphite-reinforced Al provides further evidence that the acicular particles observed in LB and GTA welds are aluminum carbide.

Evaluation of the MMC welding device

In an effort to attain more uniform properties throughout the weldment by eliminating the portion of the fusion zone in which SiC dissolution and aluminum carbide formation occurs, the MMC welding device was implemented. The final objective of the study was to determine the effectiveness of the device with respect to SiC(p)-reinforced X2080. Because the availability of the composite material was limited, “dummy” coupons were utilized in initial tests. Stainless steel and aluminum alloy test coupons were selected for preliminary testing on the basis of their respective properties which were similar to

those of the composite. While stainless steel was used as a "substitute" for the composite due to its stiffness, testing of Al alloys provided a means of predicting the relationship of thermal properties and weldability characteristics to the effectiveness of the MMC welding device.

Using the parameters established through bead-on-plate welding of 304L specimens, subsequent GTA butt welding without pressure was successfully completed on the 1 inch-wide stainless steel strips. Narrow, full penetration welds exhibiting no apparent cracking or porosity were attained at a current level of about 80 amperes and travel speed of 11 in/min, and no significant spatter, burn-through, or arc instability was observed. However, welding could not be conducted along the full length of the specimens, due to thermal expansion which generated a gap between the faying surfaces as welding progressed. The point along the length of the coupons at which gaping occurred was generally about 1 to 2 inches from the unwelded ends of the 6 to 8 inch long specimens.

The parameters corresponding to optimum results for butt welding without pressure were utilized in preliminary tests involving the application of pressure and untapered 304L specimens. Prior to the onset of welding, pressure corresponding to a regulator reading ranging from 20 to 60 psi was applied to the specimens by actuating the cylinders. Only the cylindrical straps were utilized at this point in order to restrict the upward motion of the coupons; consequently, many of the test coupons immediately buckled and deformed when subjected to lateral loading. For specimens which were not perfectly flat, the tendency for buckling to occur increased. Coupons which did not buckle or slide across one another upon contact with the rollers were subjected to butt welding with the electrode positioned

slightly in front of or behind the rollers. Regardless of the pressure applied or the position of the electrode, however, no significant effect on the quality of the weldment was observed. The same results were obtained by repeating the procedure with regulated pressure ratings up to 75 psi. Welding was completed without spatter, arc instability, or burn-through, as in previous tests, and the resulting full penetration welds appeared to be free from porosity or cracking. Due to thermal expansion as welding progressed, however, buckling and deformation of the test coupons almost invariably occurred before welding could be conducted along the full length of the specimens.

Since the hold-down straps were ineffective with respect to inhibiting buckling and/or sliding as the specimens responded to the applied load, the top tension roller was implemented for the next series of tests involving tapered coupons. Specimens having taper angles varying from 1 to 4 degrees were tested, where the best results were attained with a 1 to 2 degree taper. In these tests, the electrode was positioned no more than about 3/32" in front of the rollers, and welding was carried out without pressure along the untapered portion of the specimens, until the onset of the taper was just under the electrode. At this point, the cylinders were actuated - this rapidly advanced the drive rollers such that the load on the specimen quickly reached a maximum value corresponding to the regulator setting.

Alteration of the motor's gear ratio allowed welding to be carried out at a higher travel speed than in previous tests, in order to reduce the weld width. Due to the increased travel speed, it was necessary to use slightly higher current levels (ranging from 85 to 90 amperes) in order to achieve full penetration.

With the top tension roller in place, buckling of the coupons was avoided upon rapid actuation of the cylinders for pressure regulator settings up to about 90 psi. While advancing the drive rollers initially closed the gap between the faying surfaces, it was difficult to inhibit the eventual separation of the coupons even for high pressures. This was probably due to several factors: As shown schematically in Figure 49 [63, 64], tensile stresses ahead of and compressive stresses behind the advancing weld pool opposed the action of the rollers. Thermal expansion of the specimens also increased the difficulty in maintaining contact between the faying surfaces. The inability to overcome increasing restraint thus resulted in melting of the specimen edges rather than butt welding, as the gap between the specimens widened with progression of the slide carrier. This problem was successfully addressed, however, by placing several deep sawcuts in the outer edges of both coupons prior to welding, as shown in Figure 50. By enhancing the ability to plastically deform the material behind the weld pool, closure of the gap between the coupons was more easily accomplished. The significant thermal expansion that accompanied welding of 304L could not be entirely overcome, however, as the coupons usually buckled and deformed with sufficient heating.

In most cases, no dramatic effects on the solidification characteristics were observed for any of the procedures carried out on stainless steel samples. There was an instance, however, in which a limited amount of weld pool upset was produced at the onset of welding. Due to delayed motion of the slide carrier upon initiation of welding, arc burn-through created a "hole" in the untapered section of the test coupons. Just as slide motion resumed, the air cylinders were actuated to advance the drive rollers. The presence of the hole behind the advancing weld puddle and the rollers enabled upsetting of the joint to

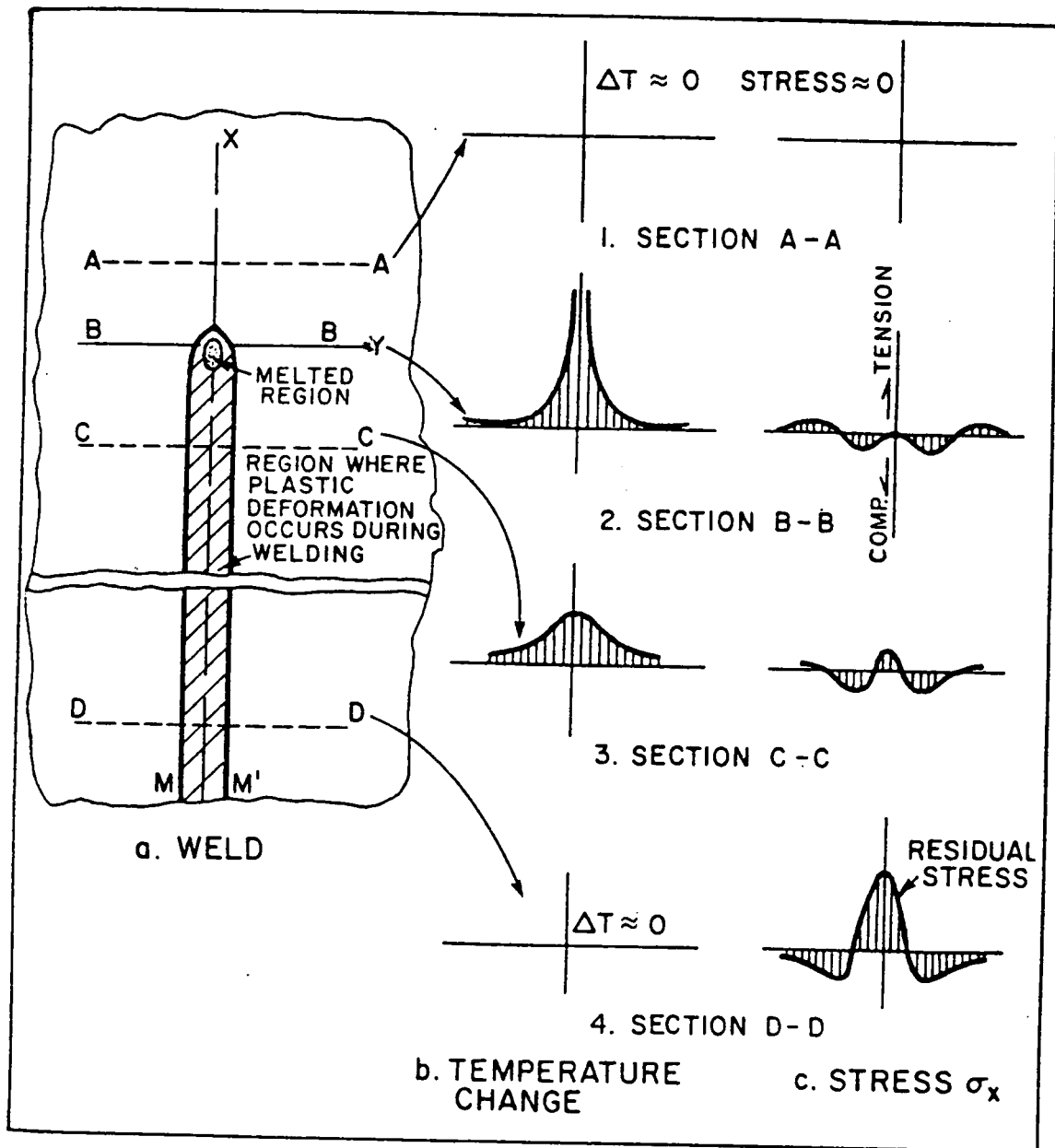


Figure 49. Schematic representation of changes in temperature and stresses during welding.[63,64]

Sources: Easterling, K., *Introduction to the Physical Metallurgy of Welding*, London: Butterworths, 1983, 34.

Masubuchi, K., Models of stresses and deformation due to welding - a review, *Modelling of Casting and Welding Processes*, Metallographic Society of AIME, 1981, 223.

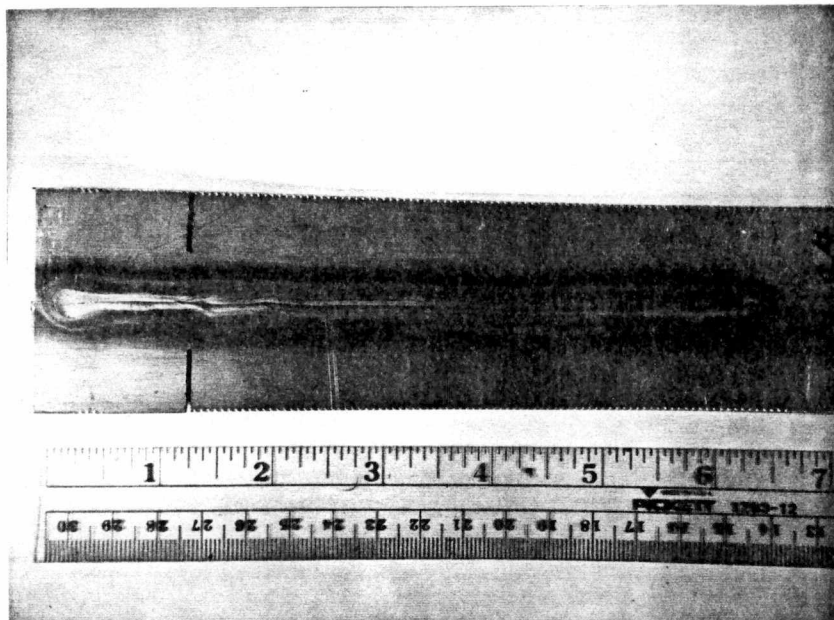


Figure 50. Stainless steel test specimen following GTA butt welding using the MMC welding assembly.

occur. As welding progressed, however, this effect was eliminated as material solidified and restraint developed behind the weld puddle.

Pressure was applied in other cases only after butt welding of the untapered section of the coupons had been completed. Although the application of pressure produced no obvious change in the "over-head" macroscopic appearance of the weldments, there was some alteration based upon examination of the underside of the coupons. This effect is shown in Figure 51, where the weld width noticeably increases with pressure relative to its original width, and retains this appearance until the welding operation ceased. Since the rollers were positioned in line with or slightly behind the electrode, the point at which widening occurred corresponded to the onset of localized deformation at the rollers' contact points (along the outer edges of the specimens). Microstructural examination of transverse weld profiles indicated that the solidification behavior was indeed altered when pressure was applied during the welding operation. In portions of the welds produced prior to the cylinder actuation, the solidification structure showed evidence of directional growth corresponding to the thermal gradients expected in butt welding, and the weld exhibited a characteristic "v-shape." The application of pressure during welding produced a fusion zone having equal widths on both the top and underside of the coupons. Although both the crown and underside of the weld were convex, there was no evidence of excessive "humping" or expulsion of molten metal from the weld zone.

Preliminary welding parameters established through bead-on-plate welding of 3003 Al alloy coupons included an alternating current level of about 70-80 amperes, 1/16" arc length, and travel speed of about 6-9 in/min. Using these parameters, butt welding was successfully conducted on untapered 2024 and 3003 coupons. Welds produced in

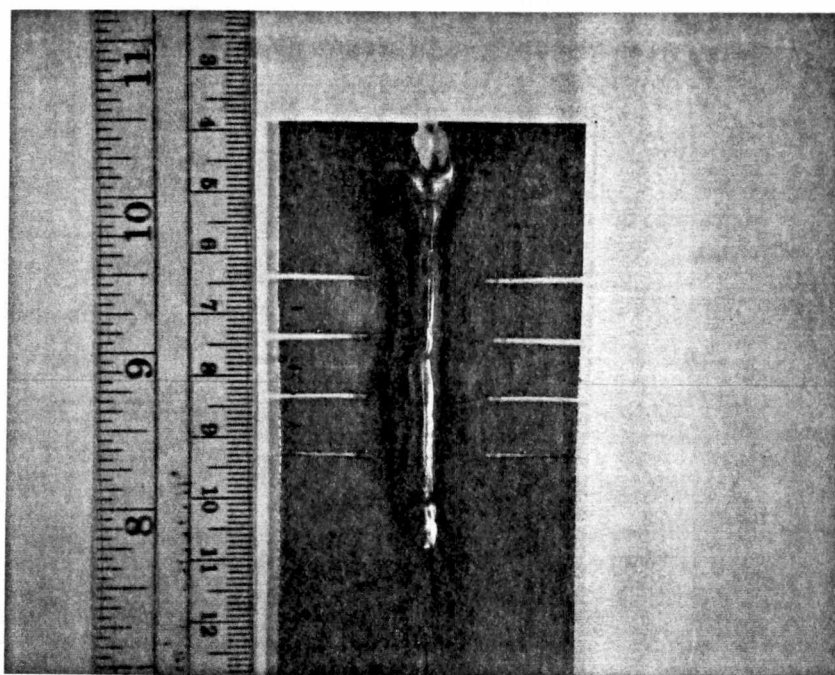


Figure 51. Butt-welded stainless steel specimen showing evidence of weld puddle alteration due to applied force during GTA welding.

specimens that were properly cleaned prior to GTAW were characterized by full penetration, no apparent cracking, and minimal (if any) porosity. Relative to the stainless steel samples, however, a significantly larger weld width corresponded to full penetration in the aluminum alloys. Additionally, the relatively low travel speed led to a progressive increase in weld width along the length of the Al alloy coupons. Attempts to reduce this width by increasing only the travel speed resulted in inadequate penetration and centerline cracking. Therefore, both the speed and current level were increased, where improved results were attained for 12 -13 in/min travel speed and 100 - 105 amperes.

Using the final set of parameters established during butt welding of the untapered Al alloy specimens, the response of tapered 3003 coupons to the application of pressure during butt welding was evaluated. The procedure previously carried out with respect to the stainless steel coupons was essentially repeated for the 3003, although the cylinder pressure was generally higher in this series of tests. When the electrode was positioned in front of the rollers, the results were similar to those obtained for the steel - a concave, full penetration weld exhibiting no evidence of expulsion or upset was produced. However, actuation of the cylinders with the rollers positioned in front of the electrode resulted in complete elimination of the gap everywhere between the faying surfaces. This occurred with every variation in the position of the rollers in front of the electrode, and for all pressure levels utilized.

Advancing the rollers in order to squeeze two "triangular-shaped" unreinforced 2024 coupons together during GTA welding in the spot mode resulted in buckling and overlapping of the specimens in every test. However, the application of this method to the SiC(p)/X2080 coupons proved to be successful, in that the desired "mechanical" objective

of utilizing the MMC welding apparatus was accomplished; that is, upsetting the joint in most cases resulted in expulsion of the fusion zone between the composite coupons.

GTA AC spot welding of the composites was initially conducted using a set of four grit-blasted coupons, including the 1/2-wide, 90 degree-tipped coupons and the triangular specimens. While partial fusion between the coupons was accomplished in some of the initial tests (based on macroscopic inspection), the use of the 1/2 inch-wide coupons did not improve the ability to join the coupons; rather, the longer period of time required in order to melt the tips of the 1/2 inch-wide coupons may have lead to drop-through of the molten pool in one test. Subsequent GTAW conducted in the spot mode with only the triangular coupons produced improved results - achieving fusion between the coupons in almost every test along with a substantial amount of upset. Further improvement in the quality of the resultant weldments was accomplished with the placement of shims on either side of the slide assembly. The approximately 1/8 inch-wide shims reduced the total amount of upset (or reduction in the total width of the welded specimen) from about 1/2 inch to 1/4 inch, and also reduced the tendency for the coupons to buckle upon upsetting with the squeeze rolls. There was, however, a slight rotation of the coupons that occurred upon upset with the squeeze rolls as a result of a minor timing difference between the two pneumatic cylinders. Since this effect could not be eliminated, the outer edges of the welded specimens were not perfectly parallel to one another. However, a reduction in the amount of buckling was achieved along with a reduction in the amount of upset, due to the addition of the shims on either side of the slide assembly.

Face and root views of a welded composite specimen that was subjected to upsetting during GTA AC spot welding is shown in Figure 52. The original width of the

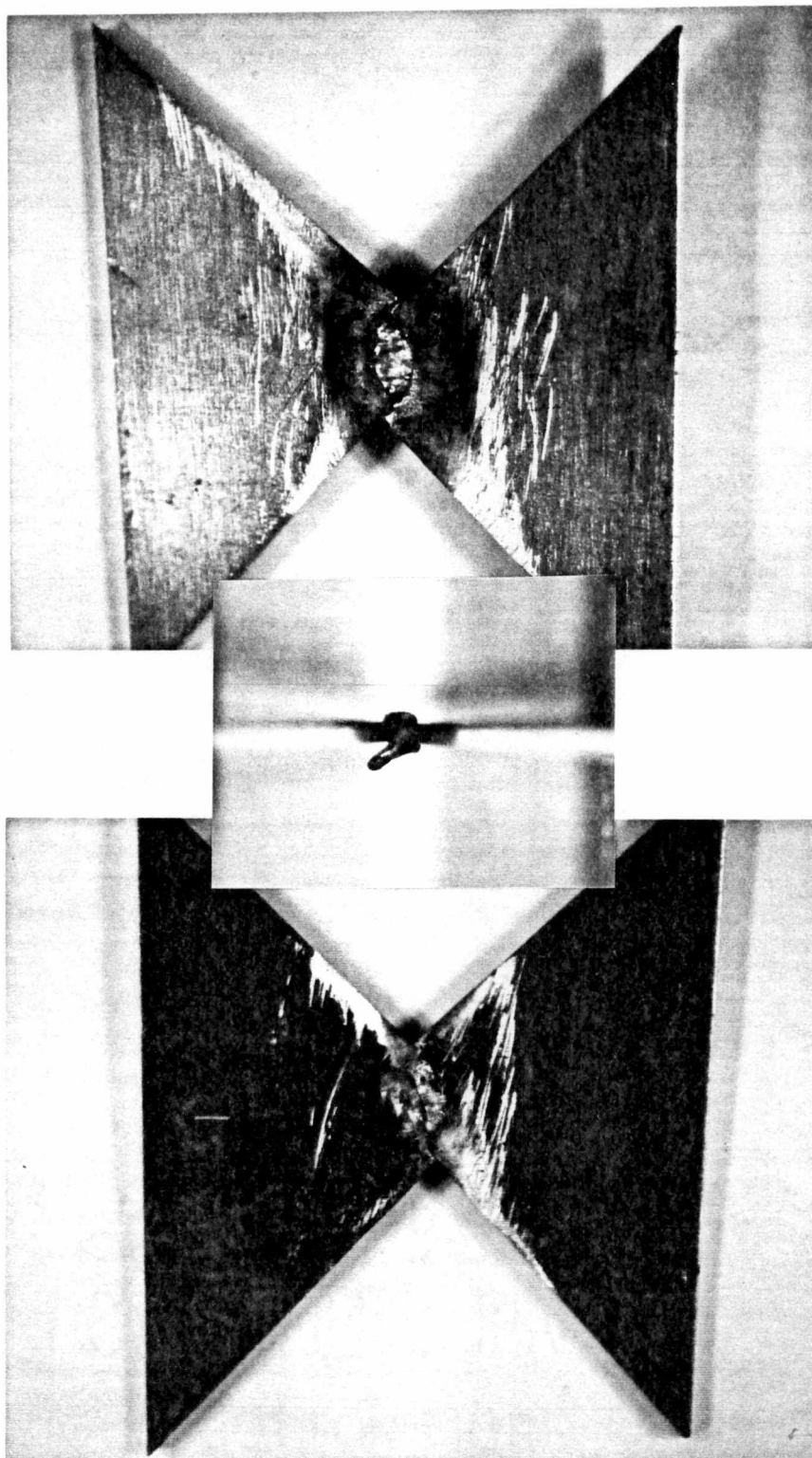


Figure 52. Face (top), profile (inset, center), and root (bottom) views of GTA spot welded SiC(p)/X2080.

abutting coupons was 2 inches in this case, and the resultant welded specimen was about 1 3/4 inches wide - corresponding to 1/4 inch upset. Based upon preliminary macroscopic inspection, it appears that there was complete bonding between the two coupons. Subsequent metallographic inspection confirms that bonding had been attained, and that a the majority of the fusion zone had been upset due to the application of force during spot welding.

Examination of a transverse profile of the weld region at low magnification revealed that the "extruded" material is characterized by a significant amount of porosity (Figure 53). The microstructural appearances at higher magnification corresponding to the upset fusion zone (a), the bond region at the center of the specimen (b), and the base material (c) are shown in Figure 54. Based upon the base material microstructure which is exhibited on either side of the bond interface between the two original coupons (region (b)), it is evident that expulsion of a majority of the fusion zone was accomplished due to the upset force. Further examination of region (b) reveals a small crack in one location at the bond interface. The microstructure in the upset region (c) is characterized by a uniform distribution of SiC particulates, an underlying cellular eutectic solidification structure, and significant amount of porosity.

Sectioning and microstructural examination of additional GTA AC spot welded specimens reveal that although significant upset occurred, complete bonding between the coupons was not obtained (Figure 55). Relative to the previous specimen examined, a larger portion of the fusion zone is retained in the joint between the coupons, rather than being extruded due to the applied pressure. Consequently, the resultant weldment is comprised of portions of a fusion zone in which higher temperature was experienced.

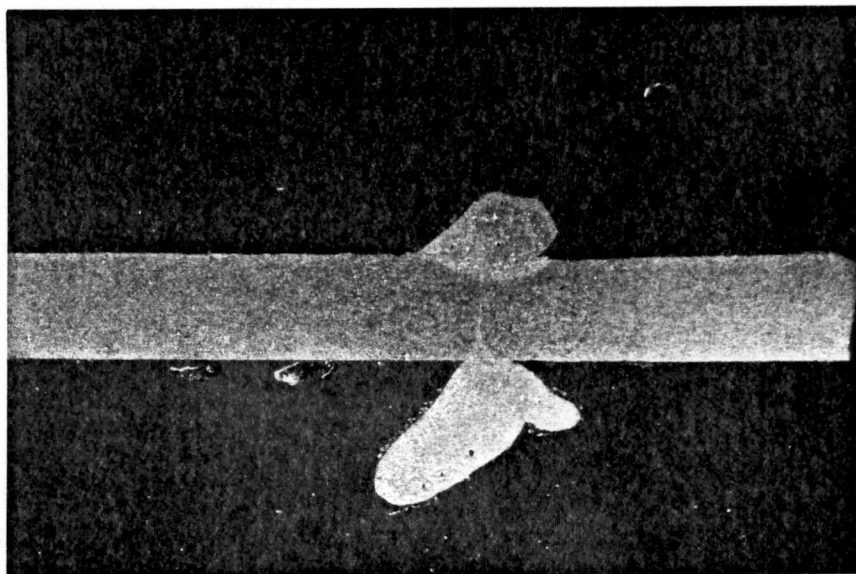


Figure 53. Cross-section of GTA spot welded SiC(p)/X2080 specimen shown in Figure 52.

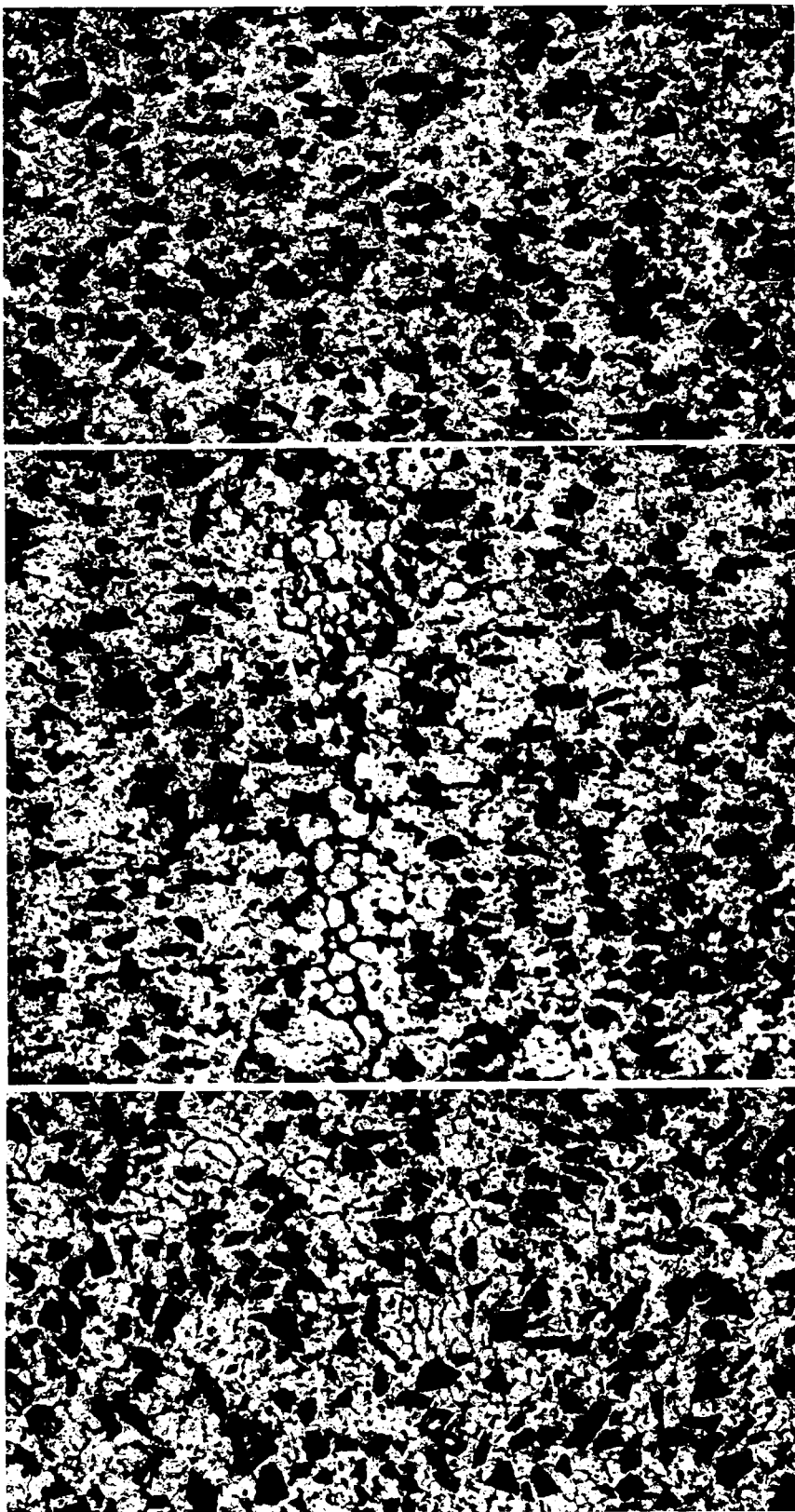


Figure 54. Microstructural appearance corresponding to the base material (right), center of specimen, and the "upset" fusion zone of specimen shown in Figures 52 and 53 (400X).

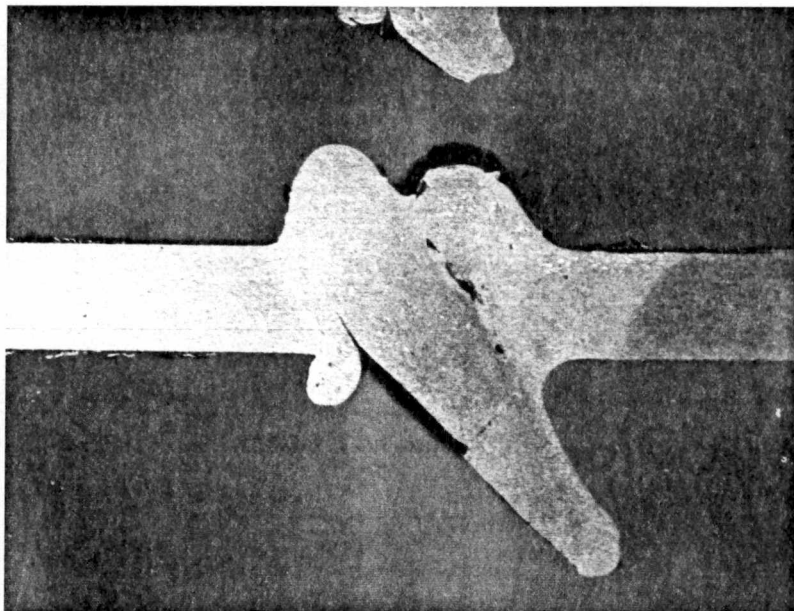


Figure 55. Cross-section of GTA-welded SiC(p)/X2080.

Figure 56 (top) shows the typical microstructural appearance of the fusion zone, in which a relatively uniform distribution of SiC particulates is retained. At the edge of a discontinuity (hole) in the center of the fusion zone shown in Figure 56 (bottom), a small number of fine, acicular aluminum carbide particles are observed. Al_4C_3 particles are also present in a small area on the edge of the "extruded" region (Figure 57). Except for these two isolated locations, however, there is minimal evidence of Al_4C_3 formation. It is likely that if the extent of fusion and upset observed in the spot weld shown in Figure 53 had been repeated in this case, the aluminum carbide would appear exclusively in the "expelled" or upset portion of the fusion zone. By additional finishing of the top and underside of the weld, the expelled or extruded material characterized by porosity and aluminum carbide formation can thus be removed to produce a weldment with improved properties.

A minimal amount of Al_4C_3 was observed in the SiC(p)/X2080 specimens that were subjected to the application of upset force during GTA AC welding in the spot mode, compared to the results for standing edge welding on the same material. This can be attributed to the rapid conduction of heat away from the weld region during spot welding, which limited high temperature exposure and inhibited SiC-Al reaction. During GTA spot welding, a stream of back purging gas was fed directly to the underside of the weld - thus providing continuous cooling action. In contrast to the standing edge-welded coupons which were positioned upright, the thin, GTA spot welded coupons were positioned such that a large surface area of the coupons was in contact with the copper chill block. The effect of the joint configuration and the addition of back purging gas was demonstrated by conducting GTA spot welding in the center of a single, triangular-shaped composite coupon without the use of back purging gas. Although welding was conducted for approximately the same period of time as in the "upset" welding tests, a significant amount

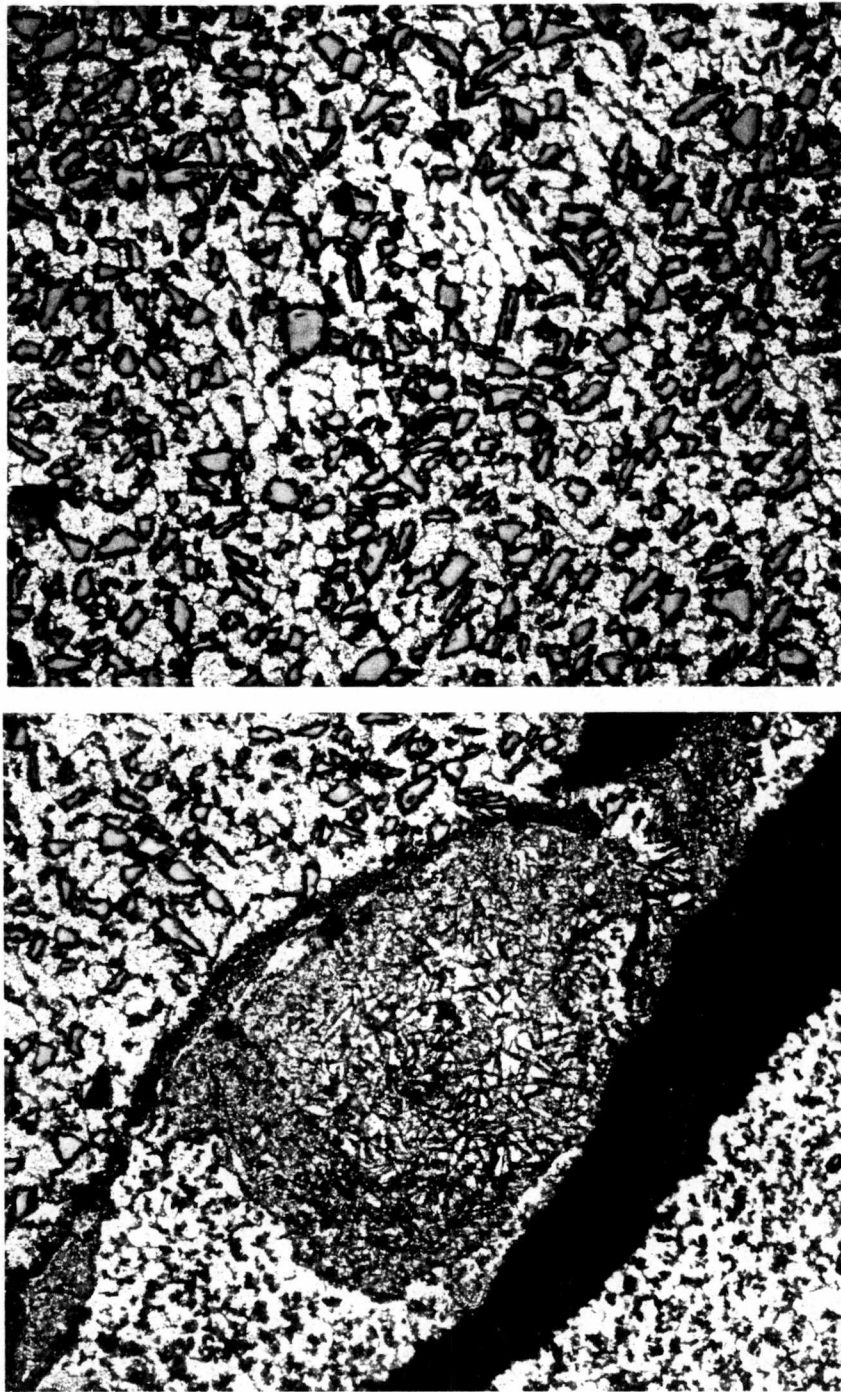


Figure 56. Fusion zone microstructure at two locations in the GTA spot welded specimen shown in Figure 55 (400X).

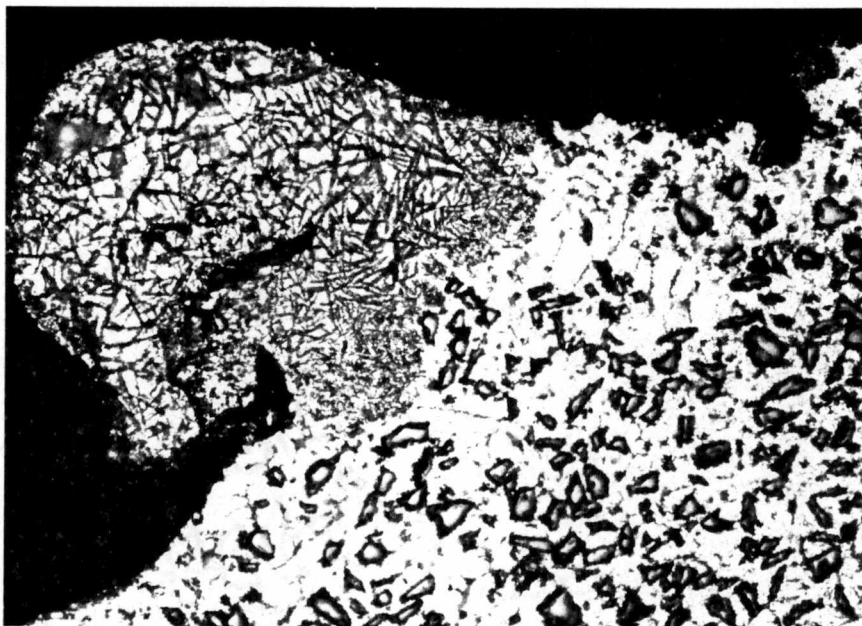


Figure 57. The Al_4C_3 constituent in the "upper" portion of the upset fusion zone in GTA spot welded $\text{SiC(p)}/\text{X2080}$ (400X).

of aluminum carbide formation was observed in the resultant weld (Figure 58). Examination of the "outer" fusion zone in this specimen revealed partial reaction of SiC, based on the "agglomerated" appearance of the particulates in this region. This indicates that high temperature exposure in this region was limited, thus resulting in retention of SiC in this portion of the fusion zone.

Bend testing

GTA AC spot welded specimens subjected to four point flexure testing were evaluated in comparison to the side-notched, edge-notched, and smooth bend specimens. The specimens were compared on the basis of the bending or rupture stress value, which was calculated according to the following equation:

$$\sigma = 3 (\text{length}) (\text{load}) / 4 (\text{width}) (\text{thickness})^2,$$

where "length" is equal to the distance between the lower bearing plate rollers, and the "load" is the maximum load applied to the specimen. The "width" and "thickness" pertain to the dimensions of the bend specimen, where the thickness was approximately 0.07 inches for all specimens.

Results of bend testing are summarized in Table 11. With the exception of one of the welded specimens (in which excessive upsetting during welding caused overlapping of the coupons to occur), the calculated stress levels are consistent for each specimen type. The lowest rupture stress values are exhibited for the side-notched specimens and the welded specimen which buckled slightly upon impact with the rollers during welding,

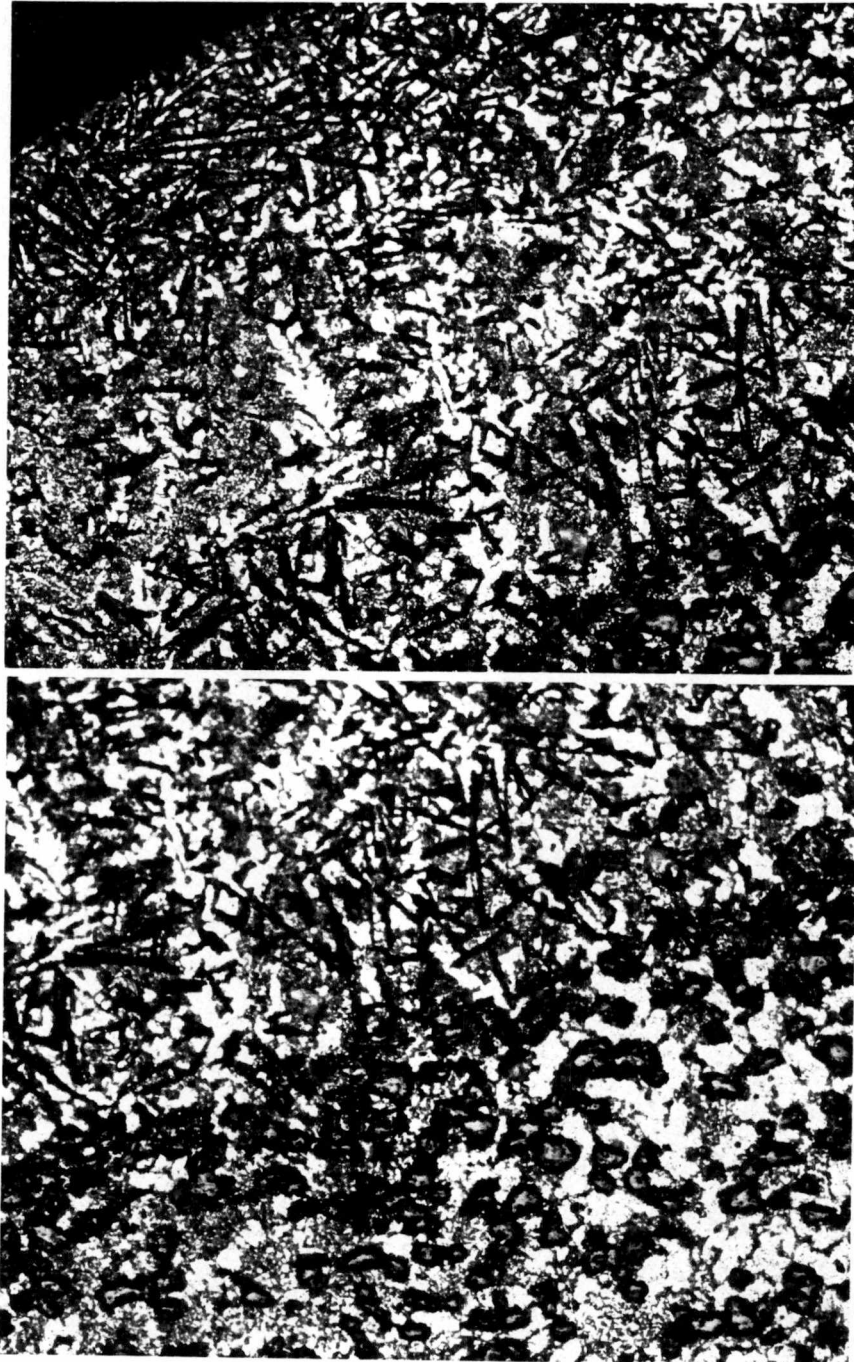


Figure 58. Fusion zone microstructure at two locations in a spot weld produced by GTAW of a single SiC(p)/X2080 coupon (400X).

TABLE 11
Results of four-point flexure tests for SiC(p)/X2080

Sample #	Specimen type	Specimen width (inches)	Cross-head speed (in/min)	Maximum load (lbs.)	Bending (σ_B) or rupture (σ_R) stress (ksi)
S1	side-notched	0.3110	0.1	59	33.6, σ_R
S2	side-notched	0.3055	0.1	54	31.3, σ_R
S3	side-notched	0.3210	0.1	60	33.1, σ_R
S4	side-notched	0.3110	0.2	60	34.1 σ_R
E1	edge-notched	0.3370	0.1	77	40.4, σ_R
E2	edge-notched	0.2970	0.1	94	56.0, σ_R
E3	edge-notched	0.2505	0.1	86	60.8, σ_R
S1	smooth	0.3270	0.1	124	67.1, σ_B
S2	smooth	0.3305	0.1	130	69.6, σ_B
S3	smooth	0.3005	0.5	105	61.9, σ_B
W1	welded	0.3110	0.1	120	68.5, σ_B
W2	welded	0.3150	0.1	120	67.4, σ_R
W3	welded	0.2915	0.1	51	31.0, σ_R

while somewhat higher values are achieved for edge-notched specimens. Since the smooth specimens and one of the welded specimens did not rupture, the calculated stress values are based on the maximum applied load rather than the breaking load. Rupture occurred in the base material of the remaining welded specimen. In general, the stresses calculated for the two bend specimens containing welds are comparable to those of the smooth specimens - indicating that positive results can be obtained for modified GTAW in the spot mode of SiC(p)/X2080.

CHAPTER V

CONCLUSIONS

The conclusions drawn from this study are summarized as follows:

1. The thermodynamic instability of SiC in contact with molten aluminum leads to formation of the compound Al_4C_3 and an Al-Si alloy above the liquidus of aluminum.
2. As the amount of silicon in liquid aluminum increases, the reaction between SiC and Al tends to saturate. By increasing the initial silicon content in the matrix alloy, the formation of Al_4C_3 can be inhibited at relatively low temperatures typically utilized in casting operations.
3. The free energy change associated with SiC-Al reaction for a constant activity of silicon becomes increasingly negative with increasing temperature. Thus, the high temperatures experienced in welding readily promote reinforcement/matrix reaction.
4. Loss of the basic engineered properties of the composite is the result of Al_4C_3 formation, due to the dissolution of the reinforcing SiC particulates and the corresponding enrichment of the matrix alloy with silicon.

5. Al_4C_3 imparts undesirable properties to the material in which it is contained, including brittle behavior and corrosion susceptibility in aqueous environments.
6. The extent of aluminum carbide formation can be inhibited by controlling the reaction kinetics through limiting high temperature exposure.
7. Temperatures encountered in conventional fusion welding processes such as GTAW readily promote the dissolution of SiC and the formation of aluminum carbide. However, the formation of the aluminum carbide is limited to a portion of the fusion zone associated with higher peak temperature thermal cycles than those encountered in the "outer" fusion zone, in which a base material distribution of SiC particulates is retained.
8. Although LBW produces lower overall energy input than GTAW, the high peak temperature thermal cycles encountered in the keyhole result in aluminum carbide formation. As in GTAW, Al_4C_3 is restricted to a portion of the LB fusion zone, while a base material SiC distribution characterizes the "outer" fusion zone corresponding to lower temperature thermal cycles. A limited "intermediate" zone located between the keyhole and "outer" fusion zone exhibits partial SiC dissolution and Al_4C_3 formation.
9. A method of mechanically eliminating a portion of the fusion zone in which SiC dissolution occurs shows promise as a means of successfully joining SiC(p)-reinforced aluminum alloys, if it can be successfully adapted to

various fusion welding techniques. The method is currently applicable to GTA spot welding of the composites, but further modification of the technique will be required to achieve improved results and to ensure repeatability.

10. The limitation of high temperature exposure during welding can result in a fusion zone characterized by a base material distribution of SiC and an absence of the Al_4C_3 constituent. This may be accomplished through the use of a joint configuration which allows rapid conduction of heat away from the weld region, such that temperatures sufficient for melting of the matrix alloy - but insufficient for SiC dissolution - are encountered in the fusion zone.

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