Capacitor Discharge Resistance Spot Welding of SiC Fiber-Reinforced Ti-6Al-4V

Tensile shear fracture occurs remote from the solid-state spot weld interface in the fiber-reinforced alloy

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ABSTRACT. Solid-state and fusion welds have been produced between sheets of monolithic and SiC fiber-reinforced Ti-6Al-4V using capacitor discharge resistance spot welding. Solid-state welds in monolithic sheet were characterized by beta grain growth across the weld interface, an alpha-prime martensite microstructure and the presence of occasional, fine interface discontinuities. Despite experiencing tensile-shear fracture along or directly adjacent to the weld interface, average tensile shear strengths for optimized solid-state welds were comparable to those of conventional fusion spot welds produced at higher energy inputs which failed by nugget pull-out. High integrity, solid-state welds were also produced in Ti-6Al-4V sheet containing 35 vol-% continuous SiC (SCS-6) fibers. Under optimized conditions, defect-free solid-state welds were produced which exhibited negligible evidence of fiber displacement or degradation. The weld zone was characterized by limited beta grain growth across the interface and a fine, martensitic microstructure. The average tensile shear strength for these welds was approximately 60% of that exhibited by optimized solid-state welds in the monolithic alloy and failed remote from the weld interface along an adjacent and parallel layer of fibers in the weld heat-affected zone.

Introduction and Background

Requirements for high-performance aircraft and aerospace systems in the 1990’s will demand the development of advanced materials which can provide physical and mechanical properties superior to those of current alloys (Ref. 1). One approach to attaining these goals has been the development of fiber-reinforced, metal-matrix composites, in which two constituent materials exhibiting significantly different physical and mechanical properties are combined in order to create a quasihomogenous material whose properties are markedly improved over those of the individual constituents (Ref. 2). Ideally, a metal-matrix composite should consist of a metallic matrix which is strong, tough and ductile, and low density, nonmetallic fibers which are extremely strong and stiff. When properly consolidated, metal-matrix composites can provide significant improvements in specific strength and stiffness over conventional monolithic alloys.

The inherent high strength and low density of titanium alloys have led to their extensive use in fiber-reinforced composites for over 20 years. During this period, the principal impediment to the widespread engineering application of these materials has been reduced tensile and fatigue properties resulting from fiber-matrix interactions during consolidation and subsequent processing. In recent years, advanced fiber coatings have been developed which reduce such interactions and allow the achievement of improved mechanical properties. These fibers, which include SCS-6 (SiC type) and B_4C/B, have been combined principally with high-strength alpha-beta and metastable-beta titanium alloys in order to provide tensile properties far superior to those of the monolithic alloys.

KEY WORDS
Metal Matrix Composites
Resistance Spot Weld
Capacitor Discharge
SiC Fiber
Tensile Shear Fracture
Solid-State Welds
Fusion Welds
Titanium
Ti-6Al-4V
Fiber Matrix Interf.
The fiber-reinforced Ti-6Al-4V composite evaluated in the present study was produced by Textron Specialty Materials, Lowell, Mass. (Ref. 2). The SCS-6 fiber utilized in this composite is produced by the chemical vapor deposition of beta SiC onto a carbon core — Fig. 1. A carbon-rich surface coating approximately 3 microns in thickness, in which the composition is graded slightly back to the SiC stoichiometry at the outer surface and midway between the inner and outer coating surfaces, is deposited during the final stage of the CVD process. This coating serves as a sacrificial layer to retard the reaction between the SiC fiber and the titanium alloy matrix during consolidation, thereby maximizing composite mechanical properties (Ref. 1). The SCS-6 fibers are approximately 5-6 mils (140 microns) in diameter and exhibit an extremely high strength (>500 ksi; 3445 MPa) and elastic modulus (60 MSI; 413 GPa), and low density (0.11 lb/in.\(^3\); 3 gm/cc). Consolidation of the sheet can be performed via either vacuum hot pressing or the hot-isostatic pressing of thin, chemically milled sheets at temperatures in the vicinity of 900° to 950°C (1652° to 1742°F). At these temperatures, the titanium flows superplastically around the fibers and bonds completely together with adjacent sheets.

Room-temperature tensile properties of monolithic and SCS-6 fiber-reinforced Ti-6Al-4V are compared in Table 1 (Refs. 1 and 3). As indicated, the composite is characterized by very high strength and stiffness in the longitudinal direction but relatively low strength transverse to the fiber direction. Despite highly anisotropic mechanical properties, these materials exhibit a strong potential for structural aerospace applications in both gas turbine-engine and airframe structures.

Aside from a high cost, the primary factor influencing the application of fiber-reinforced titanium alloys is their capability to be fabricated into structural components. In the case of monolithic titanium alloys, fusion welding has been widely utilized as an efficient and cost-effective method of fabrication. Unfortunately, the basic nature of fiber-reinforced titanium precludes the use of such techniques due to severe fiber degradation (Ref. 4). Early gas tungsten arc welding studies by Kennedy (Ref. 5) on Ti-W and Ti-graphite composites clearly showed the serious difficulties with fiber displacement and deterioration. Extensive work by Hersh (Refs. 6 and 7) in the early 1970s demonstrated the utility of resistance spot welding for the joining of continuous-fiber aluminum-boron composites. He found that through the judicious selection of welding energy input and electrode pressure, high-integrity fusion spot welds exhibiting minimal fiber degradation and displacement could be produced. Considering the appreciably higher melting point of titanium vs. aluminum, and its high chemical reactivity, a greater difficulty might be anticipated in producing similar high-integrity fusion spot welds in fiber-reinforced titanium alloys.

Solid-state welding processes, in which melting and solidification are avoided, offer an alternative to fusion welding processes for the joining of fiber-reinforced titanium alloys. Recently, diffusion welding (or bonding) has been utilized to produce SiC fiber-reinforced Ti-6Al-4V corrugated panels (Refs. 8 and 9). Typical pressures utilized in the diffusion welding of titanium alloys (0.5 to 5 MPa; 72 to 720 psi) are sufficiently low to allow welding without fiber damage and displacement. However, the thermal cycles required (which are comparable to those utilized during original consolidation of the composite) can promote further fiber/matrix degradation and property reductions. The appreciably higher levels of plastic deformation associated with other solid-state joining processes, such as friction welding, can promote significant fiber displacement and damage at the weld interface (although such a process may be very satisfactory for the joining of composite materials containing small, discontinuous fibers).

Table 1 — Room-Temperature Tensile Properties of Monolithic and SCS-6 Fiber-Reinforced Ti-6Al-4V Sheet (Refs. 1 and 2)

<table>
<thead>
<tr>
<th>Material</th>
<th>Ultimate Tensile Strength</th>
<th>Elastic Modulus</th>
<th>Elongation</th>
</tr>
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<tbody>
<tr>
<td></td>
<td>Long. Trans.</td>
<td>Long. Trans.</td>
<td>Long. %</td>
</tr>
<tr>
<td>Ti-6Al-4V/</td>
<td>890</td>
<td>890</td>
<td>120</td>
</tr>
<tr>
<td>Monolithic(a)</td>
<td>(129)</td>
<td>(129)</td>
<td>(17.5)</td>
</tr>
<tr>
<td>Ti-6Al-4V/</td>
<td>1,447-1,860</td>
<td>344-551</td>
<td>193-214</td>
</tr>
<tr>
<td>SCS-6(b)</td>
<td>(210-270)</td>
<td>(50-80)</td>
<td>(28-31)</td>
</tr>
</tbody>
</table>

(a) Mill-annealed sheet (732° C (1350°F)/2 h AC).
(b) Multi-ply, unidirectionally reinforced, 35 vol-%.

1. It should be noted that during the resistance spot welding of titanium, IR heating occurs principally within the titanium sheets (i.e., bulk heating) due to the high resistivity of titanium and not at the faying surfaces due to a high interface resistance (in contrast to steel and aluminum for which heating at the faying surfaces is much more important). Correspondingly, the location of the weld zone at the sheet faying surfaces results principally from conductive heat flow considerations, as this location is furthest from the water-cooled copper electrodes and would be expected to exhibit the highest peak temperatures.
The process is also characterized by an extremely short weld thermal cycle, typically less than 10 to 20 ms in duration. This short cycle promotes the concentration of the heat in a small zone near the weld interface, rather than allowing it to be conducted away from this region into the surrounding base metal. The process is also characterized by an extremely rapid cooling rate, which further minimizes heat-affect to the surrounding metal. In the joining of monolithic metals, energy levels are controlled such that melting is induced at the faying surfaces, thereby producing a small, highly localized fusion zone. However, if energy levels are decreased in a controlled manner, heating near the interface to high temperatures can be induced without melting. Upon the application of an appropriate electrode force, a low deformation/diffusion type weld can be produced. Titanium is well known for its ease of diffusion welding due to a high solubility for its oxide and other surface contaminants, and rapid decrease in yield strength at elevated temperatures. Consequently, such a solid-state joining process may be considered potentially ideal for the joining of fiber-reinforced titanium alloys. The principal objective of the present investigation was to examine and characterize this potential.

**Objectives**

The overall objective of the present investigation was threefold: 1) to investigate the feasibility of using CD resistance spot welding to produce solid-state welds in monolithic and fiber-reinforced Ti-6Al-4V and dissimilar welds between these materials; 2) to systematically investigate the influence of weld energy input on the nature of weld formation and resulting weld integrity and structure, with particular interest in examining the transition from solid-state to fusion welding with increasing energy input; and 3) to determine weld mechanical properties and fracture characteristics and relate these to structural characteristics.

**Experimental Procedures**

**Materials**

The fiber-reinforced Ti-6Al-4V evaluated in this investigation was produced in the form of 0.838-mm (0.033-in.) thickness sheet. As shown in Fig. 1A, the sheet contained three layers of continuous SCS-6 fibers in a cross-ply arrangement. The spacing between adjacent, parallel fibers was somewhat irregular and occasionally fibers were completely absent. Examination of the Ti-6Al-4V matrix at increased magnification revealed an equiaxed alpha grain morphology with small islands of beta phase located at alpha grain boundaries — Fig. 1B. Characterization of the fiber-matrix interface using SEM (Fig. 1C) clearly revealed the carbon-rich coating and irregular reaction zone between the coating and the titanium alloy matrix. As shown, essentially two coating layers appear to exist on the fiber. The boundary between the apparent layers (small arrow in Fig. 1C) represents the aforementioned SiC-enriched composition located approximately midway through the 3-micron-thick carbon-rich coating. Although the reaction zone product was not examined in detail, previous investigators (Ref. 10) have shown this structure to consist principally of titanium carbide (TiC) and a lesser quantity of titanium-silicide (Ti$_5$Si$_3$).

For comparative purposes, C-D resistance spot welds were also produced in monolithic Ti-6Al-4V sheet 0.762 mm (0.030 in.) in thickness. This material was provided in the alpha-beta rolled and mill-annealed condition and exhibited a typical equiaxed alpha microstructure which was noticeably finer than that observed in the fiber-reinforced product.

Coupons of 19 mm (0.75 in.) by 38 mm (1.5 in.) for CD resistance spot welding were sectioned from the as-received sheets using a thin, water-cooled abrasive cutoff wheel in order to minimize mechanical and thermal damage to the composite material. Immediately prior to welding, the coupons were chemically pickled in a solution of 5 mL HF + 40 mL HNO$_3$ + 55 mL H$_2$O, rinsed with distilled water, methanol and dried. Profilometer measurements (in terms of R$_a$ number) determined the surface roughness of the fiber-reinforced sheet to be greater than that of the monolithic sheet (1.8 vs. 0.6 micrometers).

**Capacitor Discharge Resistance Spot Welding**

Spot welds were produced in both sheet materials using a Kimura-Denyoki Model SC-40 capacitor discharge resistance welding system. The capacitor discharge spot welding sequence is schematically illustrated in Fig. 2 (Ref. 11). As shown, the sheet coupons first experience a squeeze force, after which the force increases to the welding force. On reaching the welding force, the welding current is initiated via the discharge of a capacitor bank preset at a specific voltage. The force continues to increase to a forging force, which is maintained until completion of the welding cycle. The peak current value and energy input into the weld are controlled principally by the capacitor voltage (assuming a constant total capacitance), while the duration of current flow is dependent on the reactance of the welding system and across the weld. It should be noted that the principal variable in the process is the capacitor voltage, as this value determines the total energy input into the weld.

Based on literature provided by Kimura-Denyoki (Ref. 11) and preliminary welding trials, a satisfactory electrode diameter and tip geometry, and an acceptable range of welding voltages and electrode forces were determined which provided satisfactory weld quality while minimizing the displacement and damage to fibers in the composite. RWMA Class 2 copper electrodes 6.35 mm (0.25 in.) in diameter and exhibiting a 45-deg angle truncated cone tip geometry were dressed to produce a flat contact surface and cleaned with methanol immediately prior to welding. Maintenance of a constant electrode alignment and diameter was verified throughout the test program by taking imprints of the contacting electrodes with carbon paper. Based on the preliminary analysis, final iteration welds for
detailed metallographic characterization and mechanical property analysis were produced using a constant electrode force of 7.12 kN (1600 lb) and a range of capacitor voltages from 150 to 180 V for the monolithic material, 170 to 190 V for the fiber-reinforced material and 160 to 180 V for the dissimilar material welds. Through this systematic variation in energy input, a continuous range of weld types from completely solid state to fusion were generated. The limited quantity of fiber-reinforced material required multiple parameter variations (i.e., voltage and fiber orientation) for each set of test specimens. Welds produced in the fiber-reinforced material at 170 and 190 V and between the monolithic and composite material were generated with the outer fibers of both coupons oriented perpendicular to the longitudinal coupon axis. In contrast, welds produced in the fiber-reinforced sheet at 180 V were generated with the outer fibers of the two coupons oriented perpendicular to each other. Generally, four welds were produced for each combination of welding conditions. One specimen was utilized for metallographic characterization and the remaining three specimens were tension shear tested.

Weld Characterization

Samples for microstructure characterization were obtained by sectioning the spot welds transversely through the center of the weld zone using a low-speed diamond saw, usually in a direction parallel to the longitudinal axis of the welded coupon. Specimens were subsequently mounted in epoxy, ground down to 600-grit SiC and polished down to 9 microns using a diamond compound. Following final polishing with 0.06 μm colloidal silica, samples were etched with Kroff’s reagent (2 mL HF + 6 mL HNO₃ + 92 mL H₂O). Characterization of the weld structures included both conventional light microscopy and scanning-electron microscopy, and energy-dispersive x-ray analysis.

Microhardness traverses were performed on representative welds using a Knoop microhardness tester with a 500-g load. Through-thickness hardness traverses at the weld centerline were performed on each weld type. In addition, hardness traverses parallel to the weld interface were performed 0.06 mm (0.0023 in.) from the weld interface.

Tensile-shear testing was performed on samples 76 mm (3.0 in.) in length and 19.5 mm (0.75 in.) in width. Samples were clamped 12.7 mm (0.5 in.) on each side of the weld, and the specimen was subjected to a tensile load. The specimens were then transferred to the shear testing machine, and the shear load was applied. The tensile load was increased until the specimen failed, and the shear load was recorded. The tensile-shear strength was determined as the maximum load divided by the area of the weld interface. The tensile-shear stress was calculated by dividing the tensile-shear strength by the weld interface area.

Table 2 — Welding Parameters and Mechanical Properties for Capacitor-Discharge Resistance-Spot Welds Produced in Monolithic and Fiber-Reinforced Ti-6Al-4V(4)

<table>
<thead>
<tr>
<th>Specimen Type/Number</th>
<th>Voltage V</th>
<th>Peak Current A</th>
<th>Weld Type((c))</th>
<th>Central Interface Hardness</th>
<th>Weld Diameter(mm (in.))</th>
<th>Tensile-Shear Strength (nms)</th>
<th>Tensile-Shear Stress(MPa (ksi))</th>
<th>Fracture Mode((e))</th>
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<tr>
<td>Ti-6-M-1</td>
<td>150</td>
<td>5.300</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>4.12 (926)</td>
<td>259 (37.6)</td>
<td>IS</td>
</tr>
<tr>
<td>Ti-6-M-2</td>
<td>150</td>
<td>5.500</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>4.24 (953)</td>
<td>267 (38.8)</td>
<td>IS</td>
</tr>
<tr>
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<td>150</td>
<td>5.800</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>5.77 (1.300)</td>
<td>364 (52.8)</td>
<td>IS</td>
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<td>—</td>
<td>4.71 (1.060)</td>
<td>297 (43.1)</td>
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<td>5.525</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>5.49 (1.235)</td>
<td>327 (47.5)</td>
<td>IS</td>
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<tr>
<td>Ti-6-M-1</td>
<td>160</td>
<td>5.900</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>5.22 (1.175)</td>
<td>311 (45.2)</td>
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<tr>
<td>Ti-6-M-2</td>
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<td>5.700</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>6.11 (1.375)</td>
<td>364 (52.9)</td>
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<td>6.030</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>5.60 (1.261)</td>
<td>334 (48.5)</td>
<td>IS</td>
</tr>
<tr>
<td>Ti-6-M-1</td>
<td>170</td>
<td>6.300</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>5.66 (1.275)</td>
<td>313 (45.5)</td>
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<td>6.100</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>5.22 (1.175)</td>
<td>289 (41.9)</td>
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<tr>
<td>Ti-6-M-Avg</td>
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<td>6.250</td>
<td>FW</td>
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<td>—</td>
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<td>Ti-6-M-2</td>
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<td>6.800</td>
<td>FW</td>
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<td>—</td>
<td>5.84 (1.315)</td>
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<td>6.800</td>
<td>FW</td>
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<td>—</td>
<td>5.60 (1.261)</td>
<td>334 (48.5)</td>
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<tr>
<td>Ti-6-M-1</td>
<td>180</td>
<td>6.800</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>5.55 (1.230)</td>
<td>274 (39.8)</td>
<td>IS</td>
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<tr>
<td>Ti-6-M-2</td>
<td>180</td>
<td>6.700</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>5.50 (1.275)</td>
<td>279 (40.6)</td>
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<td>—</td>
<td>5.55 (1.230)</td>
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<td>Ti-6-M-1</td>
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<td>4.000</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>3.44 (775)</td>
<td>267 (38.7)</td>
<td>FMI</td>
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<td>3.600</td>
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<td>3.48 (785)</td>
<td>270 (39.2)</td>
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<td>3.900</td>
<td>SS</td>
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<td>3.46 (775)</td>
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<td>FMI</td>
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<td>3.833</td>
<td>SS</td>
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<td>—</td>
<td>3.37 (760)</td>
<td>189 (27.4)</td>
<td>FMI</td>
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<td>4.100</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>3.55 (800)</td>
<td>199 (28.8)</td>
<td>OP</td>
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<td>Ti-6-M-Avg</td>
<td>180</td>
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<td>SS</td>
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<td>4.300</td>
<td>SS</td>
<td>—</td>
<td>—</td>
<td>3.56 (800)</td>
<td>199 (28.9)</td>
<td>FMI or OP</td>
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<tr>
<td>Ti-6-M-2</td>
<td>180</td>
<td>4.000</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>3.95 (890)</td>
<td>271 (39.4)</td>
<td>FMI</td>
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<tr>
<td>Ti-6-M-Avg</td>
<td>180</td>
<td>4.300</td>
<td>FW</td>
<td>—</td>
<td>—</td>
<td>3.95 (890)</td>
<td>271 (39.4)</td>
<td>FMI</td>
</tr>
</tbody>
</table>

Table 2 — Welding Parameters and Mechanical Properties for Capacitor-Discharge Resistance-Spot Welds Produced in Monolithic and Fiber-Reinforced Ti-6Al-4V(4)

(a) Average of all specimens tested per condition. Electrode force was constant at 7.12 kN (1600 lb) for all welds.

(b) Monolithic, MMC-metal-matrix composite.

(c) Solid-state weld, FW-fusion weld.

(d) Hardness of central interface region in MMC; hardness of central interface region in monolithic alloy.

(e) Measured from metallographic cross-section.

(f) Based on radius measured from metallographic cross-section.

(g) Interface shear, NP-nugget pullout, FMI-fiber-matrix interface.
end and tested at an extension rate of 0.134 mm/s (0.005 in./s). Fracture shear stress was calculated from the fracture strengths and metallographic measurements of the weld cross-sectional areas. Since these specimens do not experience pure shear during testing, values of shear stress at fracture must be used with caution. In the present work, these values were calculated for the purpose of allowing a comparison of weld strengths normalized to account for differences in the weld cross-sectional area.

Analysis of the as-fractured specimens was performed using both light and scanning-electron microscopy. In addition, selected specimens were precision sectioned using a slow-speed diamond saw, mounted, polished and etched in order to more accurately identify the fracture path and mechanism.

Results

Welding Process Performance

Peak current values recorded for each material combination and capacitor voltage level are listed in Table 2, and show only marginal variations for identical welding conditions (typically less than 5% from the average). For identical capacitor voltage levels, the peak welding currents in the monolithic Ti-6Al-4V were consistently lower than those for the SiC fiber-reinforced material (e.g., at 170 V the average peak currents were 6250 and 3833 A for the monolithic and SiC fiber-reinforced materials, respectively). Considering Ohm’s law, this difference in peak current indicated a higher resistance across the weld in the fiber-reinforced material. A higher total resistance both across the sheet/electrode and sheet/sheet interfaces would be expected due to the greater measured surface roughness of the fiber-reinforced material. As indicated previously, however, for the resistance spot welding of titanium heating is provided principally by bulk resistance heating. The higher bulk electrical resistance of the SiC fiber-reinforced material, due to the 35 volume percent carbon and the presence of fiber-matrix interfaces, therefore, was the principal contributor to the higher total resistance across this material. The peak current across the dissimilar monolithic/fiber-reinforced Ti-6Al-4V spot welds was intermediate between the similar material welds. A comparison of cross-sections of welds produced at identical voltages showed interface melting to initiate at 170 V for the monolithic material but not until 190 V for the fiber-reinforced material. Considering the aforementioned differences in electrical resistance

Fig. 3 — Solid-state CD resistance spot weld produced between Ti-6Al-4V sheets at 160 V. A — Light macrograph of cross-section through weld zone; B, C — micrograph of weld interface (arrows) at outer periphery; D, E — micrograph of weld interface (arrows) at axial centerline.

Weld Macrostructure Analysis

Capacitor discharge resistance spot welds characterized in this investigation were produced at relatively low energy inputs as compared to conventional resistance spot welds in titanium. As a result, defects commonly associated with excess heat input, such as the expulsion of molten metal from the fusion zone, or excess indentation of the sheet surfaces by the copper electrodes, were not observed. The use of lower than normal energy inputs in order to develop entirely solid-state welds correspondingly resulted in smaller weld cross-sections than typically observed in optimized fusion spot welds. Despite this reduction, all welds met the minimum diameter requirements as specified in Mil-Spec MIL-W-6838D (Ref. 12). (Note that this specification precisely applies only to fusion spot welds and that electrode geometries utilized in this study did not conform with those specified.)

Weld Microstructure Analysis

Monolithic Ti-6Al-4V Welds

Figure 3A shows the cross-section through a solid-state weld produced between monolithic Ti-6Al-4V sheets using the CD resistance spot welding process at 160 V (5850 A average peak current). At the weld outer periphery, the microstructure appeared similar to the
equiaxed alpha + beta microstructure exhibited by the unaffected base metal — Fig. 3B. Examination of this region at increased magnification (Fig. 3C), however, revealed that it had actually experienced temperatures above the beta transus (that temperature at which the low-temperature HCP alpha phase transforms completely to the high-temperature BCC beta phase on heating) during the weld thermal cycle. The thermal excursion of this region to peak temperatures only marginally above the beta transus and for an extremely short period of time allowed only minimal beta grain growth and negligible homogenization of Al and V, which were originally partitioned to the alpha and beta phases, respectively. On subsequent rapid cooling, Al-enriched, V-depleted beta phase regions, which were originally equiaxed alpha grains, transformed to fine, acicular alpha-prime (HCP) martensite, as shown in Fig. 3C, while the V-enriched beta phase regions were likely retained as beta phase. Evidence of a distinct weld interface in this region suggested incomplete metallurgical bonding and only limited beta grain growth across the interface. Nearer to the center of the weld, peak temperatures were appreciably higher, which promoted extensive beta grain growth across the interface and the complete homogenization of alloying elements in the high-temperature beta phase field — Fig. 3D. The extremely rapid cooling rates experienced by this region promoted transformation of the high-temperature beta phase to alpha-prime martensite. Despite the high temperatures experienced, and beta grain growth across the interface, examination of the weld interface at increased magnification revealed the occasional presence of small discontinuities. Structural characteristics of solid-state welds produced at 150 V (5425 A average peak current) essentially paralleled those described above, but were contained in a smaller overall weld zone.

An increase in capacitor voltage to 170 and 180 V (and correspondingly in the weld energy input) promoted appreciable melting at the monolithic sheet interfaces and the formation of a fusion zone. As shown in Fig. 4A, for a weld produced at 170 V (6250 A average peak current), outside of the fusion zone a region of solid-state welding existed where the microstructure ranged from equiaxed alpha at the weld outer periphery to a coarsened prior-beta grain/alpha-prime martensite structure near the fusion line — Fig. 4B and C. Interestingly, in contrast to the solid-state welds described above, no evidence of interface discontinuities was observed in the solid-state regions of the fusion welds. The weld fusion zone solidified epitaxially from beta grains at the fusion line. These columnar grains grew in a direction parallel to the maximum temperature gradient at the solid-liquid interface, which was parallel to the electrode axis, and ultimately impinged at the weld centerline to create an irregular beta grain boundary — Fig. 4D. As in the solid-state welds, rapid cooling rates promoted beta decomposition in the fusion zone to alpha-prime martensite.

SiC Fiber-Reinforced Ti-6Al-4V Welds

Figure 5 shows cross-sections through solid-state (A, B) and fusion (C) welds produced in the SiC fiber-reinforced Ti-6Al-4V sheets using the CD resistance spot welding process at 170, 180 and
from that observed in the monolithic alloy. Microstructural analysis clearly revealed a cellular-dendritic solidification structure, with evidence of second phase formation in the last-to-solidify interdendritic regions — Fig. 6D. A transformed-beta microstructure was not observable optically, possibly due to preferential etching of the interdendritic regions.

Since mechanical property degradation in fiber-reinforced titanium alloys is dependent to a great extent on the nature of the fiber-matrix interactions, the fiber-matrix interfaces were examined for welds produced at each energy level using scanning electron microscopy. Figure 8A and B shows the weld interface (arrow) and fiber-matrix interface at high magnification. Although the region adjacent to the fiber had experienced temperatures exceeding the beta transus, a comparison with Fig. 1C indicates that no apparent degradation of the carbon coating or increase in the size of the reaction zone occurred during the rapid weld thermal cycle. Localized melting around fibers in the weld produced at 180 V (Fig. 8C and D) promoted appreciable degradation of the carbon coating, nearly completely eroding the outer coating layer. The irregularly shaped reaction zone between the remaining coating and matrix was somewhat larger than that of the unaffected base metal. The melted and resolidified microstructure adjacent to the fiber consisted of a coarse solidification structure, which contained an acicular appearing interdendritic phase. Since the titanium-rich side of both the Ti-Si and Ti-C binary phase diagrams exhibit a eutectic (i.e., $k_1 < 1$), it is suggested that the solid-state diffusion of C or Si from the fiber coating into the surrounding matrix during the weld thermal cycle depressed its solidus temperature, thereby promoting localized melting. Considering these eutectic diagrams, the formation of either titanium-carbide (TiC) or a titanium-silicide ($\text{Ti}_2\text{Si}_3$) is indicated in the interdendritic regions. Energy-dispersive x-ray spectroscopy (EDS) analysis of the acicular phase did not indicate appreciable Si enrichment, suggesting the phase to be titanium carbide (carbon could not be detected with EDS analysis). This observation was consistent with erosion of only the outer, carbon-rich coating rather than the SiC fiber interior. Figure 8E and F shows the fiber-matrix interface within the fusion zone of a
Fig. 7 — SEM micrographs of weld interface at axial centerline of solid-state CD resistance spot weld produced between SiC fiber-reinforced Ti-6Al-4V sheets at 160 V. A — secondary electron image; B — back-scattered electron image.

Fig. 8 — SEM micrographs showing fiber-matrix interfaces adjacent to weld interfaces adjacent to weld interface for CD resistance spot welds produced between SiC fiber-reinforced Ti-6Al-4V: A, B — 170 V; C, D — 180 V; E, F — 190 V.

Fig. 9 — Light macrographs of cross-section through solid-state CD resistance spot welds produced between monolithic and SiC fiber-reinforced Ti-6Al-4V sheets. Left — 160 V; right — 170 V.
weld produced at 190 V. Although appreciably eroded, evidence of the remaining carbon coating, and an irregu-
larly shaped reaction zone, can still be observed. Assuming that a liquid/carbon coating interface existed at peak tem-
peratures during the weld thermal cycle, it is apparent that the reaction zones ob-
served in both Figs. 8D and 8F likely formed during the on-cooling portion of the weld thermal cycle.

**Dissimilar Monolithic/SiC Fiber-Reinforced Ti-6Al-4V Welds**

Figure 9 shows light macrographs of cross-sections through solid-state CD resistance spot welds produced between monolithic and fiber-reinforced Ti-6Al-
4V sheets. The weld produced at 160 V showed complete solid-state welding at the weld interface with no evidence of melting. Although the weld produced at 170 V also exhibited only solid-state welding at the weld interface, extensive melting was observed in the fiber-rein-
forced sheet.

Microstructural characteristics of the weld interface at the outer periphery and axial centerline paralleled those ob-
served for the similar alloy welds between the monolithic and fiber-rein-
forced sheets — Figs. 10A-D.

**Mechanical Property Analysis**

**Hardness Testing**

Hardness traverses across solid-state spot welds produced in monolithic Ti-
6Al-4V showed an increase from approximately 320 KHN in the unaffected base metal to a maximum of 375 KHN at the center of the weld interface — Table 2. This increase in hardness was consistent with the transition from the equiaxed alpha + beta base metal microstructure to the martensitic alpha-
prime martensite structure at the center of the weld interface. Hardness traverses across fusion spot welds in Ti-6Al-4V showed a similar increase in hardness, but exhibited a slightly higher maximum hardness within the fusion zone of 385 KHN. Considering the close similarities in the martensitic microstructures exhibited at the centers of both solid-state and fusion welds, the origin of this hardness difference was not clear.

Hardness traverses performed across solid-state welds produced in the SiC fiber-reinforced material indicated an increase in hardness from approximately 325 KHN in the unaffected base metal to 390 to 400 KHN at the outer periphery of the weld interface, to a maximum of 410 to 425 KHN at the center of the weld interface. Hardness values measured across the fiber-reinforced mate-
rial generally exhibited greater fluctuations as compared to the monolithic ma-
terial. The greater hardness associated with the alpha-prime martensite microstructure observed in this region, vs. a comparable microstructure observed in the monolithic sheet welds, was attributed to compositional variations, particularly a higher interstitial level in the fiber-reinforced materials. Micro-
hardness evaluation using a low inden-
tor load of 100 g showed that hardness of the titanium matrix was not influenced by distance from the fibers, except at locations directly adjacent to fibers where localized melting had occurred. Hard-
ness levels in these resolidified regions up to 450 KHN were measured. Fusion welds in the fiber-reinforced alloy exhibited a significant increase in hardness vs. the solid-state weld interface, to a maximum of 460 to 485 KHN. The increase in hardness was attributed to substitutional and interstitial solid solution strengthening from Si and C, respectively, a change in the type or mor-
phology of the martensitic microstruc-
ture present in the rapidly cooled fusion zone, and the effect of fine carbide or silicide second-phase particles in the martensitic matrix.

**Fig. 10** — Light micrographs of solid-state CD resistance spot welds produced between SiC fiber-reinforced (top) and monolithic (bottom) Ti-6Al-4V sheets. A, B — 160 V; C, D — 170 V. A, C — weld interface (arrow) at weld outer periphery; B, D — weld inter-
face (arrow) at axial centerline.
Fig. 11 — As-fractured tensile-shear specimens produced from CD resistance spot welds between SiC fiber-reinforced Ti-6Al-4V sheets. A — 170 V; B, C — 180 V. Loading direction is vertical.

As-fractured tensile-shear specimens produced from CD resistance spot welds between SiC fiber-reinforced Ti-6Al-4V sheets. A — 170 V; B, C — 180 V. Loading direction is vertical.

Hardness traverses across dissimilar material weld zones directly adjacent to the weld interface paralleled those described above for the similar material welds. Maximum hardness values at the interface of solid-state welds were about 380 to 390 KHN and 405 to 425 KHN on the monolithic and fiber-reinforced material sides, respectively. Melted and resolidified locations in the weld produced at 170 V remote from the interface exhibited high hardness up to 460 KHN.

Tensile-Shear Testing

Results of tensile-shear testing are shown in Table 2. Except for welds produced in the monolithic alloy at 150 V, reproducibility of the tensile-shear test results was excellent. As shown, the tensile-shear strengths of solid-state welds produced in the monolithic alloy at 150 V were lower than those exhibited by the fusion spot welds. In contrast, the average tensile-shear strength of the solid-state weld produced at 160 V (5.60 kN; 1261 lbs) was comparable to the fusion spot welds produced at 170 V (5.47 kN; 1230 lb) and 180 V (5.68 kN; 1280 lb). These values meet minimum requirements as set forth by MIL-Spec. MIL-W-6858D for fusion welds in titanium sheet of this thickness (4.2 kN; 950 lb minimum, 5.48 kN; 1235 lb minimum average) (Ref. 12). It is important to recognize that the fusion spot welds produced and evaluated in this study were not optimized from a mechanical property standpoint. It is anticipated that further increases in weld energy input and/or electrode diameter, and correspondingly the generation of a fusion zone of greater diameter, could provide higher tensile-shear strengths than obtained in this study. As shown in Table 2, the average tensile shear stress for the solid-state welds produced at 160 V was superior to that of the fusion welds, and quite comparable to that of conventional fusion spot welds in Ti-6Al-4V sheet (Ref. 13).

Tensile-shear strengths and stresses measured for welds produced in the SiC fiber-reinforced material were consistently below those for the monolithic material. Average tensile-shear strength of 3.46 kN (778 lb) and 3.56 kN (800 lb) for solid-state welds produced at 170 and 180 V, respectively, were approximately 65% of the strength of optimized solid-state welds produced in the monolithic alloy. Tensile-shear stress levels of the solid-state weld produced at 170 V were approximately 90% of the optimized solid-state weld in the monolithic alloy.

Tensile-shear strengths of dissimilar-material welds were greater than similar-material welds produced in the fiber-reinforced alloy due to the somewhat greater cross-sectional area of the dissimilar-material welds, while the tensile-shear strengths of the similar and dissimilar-alloy welds containing the fiber-reinforced material were nearly identical.

Fracture Analysis

Fig. 12 — Light macrographs of cross-sections through tensile-shear specimens produced from C-D resistance spot welds between SiC fiber-reinforced Ti-6Al-4V sheets. Left — Weld produced at 180 V that fractured along fiber matrix interface; right — weld produced at 180 V that fractured principally through unaffected sheet. Arrows indicate weld interface.

Failure of the solid-state CD resistance spot welds produced in monolithic Ti-6Al-4V occurred macroscopically by shear fracture through the weld region. Examination of the fracture path by metallographic cross-sectioning revealed that fracture followed the weld interface only at the outer weld periphery, perhaps due to the presence of occasional discontinuities at the interface. Near the...
center of the weld zone, the fracture propagated away from the interface into the surrounding HAZ. Failure of fusion welds in the monolithic alloy occurred by fracture around the outer periphery of the weld fusion zone by nugget pullout.

Fracture of solid-state CD resistance spot welds produced in the fiber-reinforced sheet occurred in two different modes, as shown in Fig. 11. All welds produced at 170 V (Fig. 11A), and one weld produced at 180 V (Fig. 11B) failed along a layer of fibers directly adjacent to the weld interface in Fig. 12A. In welds produced at 170 V, specimens were oriented such that the layer of fibers adjacent to the weld interface were always oriented perpendicular to the applied load, as indicated in Fig. 11A. Welds produced at 160 V, however, were produced with the outer fibers for the two sheets oriented perpendicular to each other. Interestingly, the failure of a specimen parallel to the weld interface occurred along a layer of fibers oriented parallel to the load direction. Fracture of two welds produced at 180 V occurred by apparent nucleation in the vicinity of the weld outer periphery, but propagation through the base metal, as shown in Fig. 11C.

Fracture paths for the two failure modes were examined in greater detail by sectioning through the failed specimens, as shown in Fig. 12A and B. Figure 12A, which illustrates a weld produced at 170 V, shows that fracture initiated at or near the notch present between the two sheets at the outer periphery of the weld interface, and propagated along a layer of fibers parallel to the interface. Note that fracture propagated completely along the top layer of fibers, and nearly one-half the distance across the weld on the opposite layer of fibers. Figure 12B, which is a cross-section through the specimen shown in Fig. 11C, indicates crack initiation in the vicinity of the weld outer periphery notch but propagation through the base sheet.

SEM fractographic analysis was performed to identify more precisely the fracture path through welds in the fiber-reinforced alloy. Figure 13 shows the fracture surface of the tensile shear specimen which failed along a layer of fibers, and indicates delamination-type fracture between the carbon coating and the matrix, or possibly between the coating and the fiber. The titanium alloy ligaments between the fibers failed by ductile, mechanical fracture. Fractographic analysis of the specimen, which failed transversely across the sheet (Fig. 14), also revealed evidence of delamination between the fiber coating and the fiber interior and titanium matrix.

Fracture of dissimilar material CD resistance spot welds between the monolithic and fiber-reinforced Ti-6Al-4V sheets occurred in a manner identical to that observed in the solid-state welds produced between the fiber-reinforced sheets at 170 V.
Discussion of Results

Application of the CD Resistance Spot Welding Process

Results of this study have shown that the CD resistance spot welding process is effective in producing high-integrity spot welds between sheets of monolithic and fiber-reinforced Ti-6Al-4V. As indicated in the Introduction, utilization of the CD vs. conventional resistance spot welding was originally made based on the desire to concentrate heat in a highly localized weld zone near the sheet surfaces and to minimize the weld thermal cycle in the surrounding weld zone such that degradation of the fiber and fiber-matrix interface would not occur. As shown in Figs. 5 and 6, using optimum CD resistance spot welding conditions a defect-free solid-state weld could be produced in which fiber degradation was precluded. Although a distinct advantage in minimizing fiber degradation, it is apparent that a potential disadvantage of the rapid thermal cycle promoted by the CD process may be the formation of a high strength, but low ductility martensitic weld structure during rapid weld cooling. The observation that fracture occurs exclusively through the HAZ and unaffected base metal suggests that the hard interface structure effectively transfers load to these softer, surrounding regions and therefore that it is not critical to tensile shear properties of the composite weld zone.

Conventional resistance spot welding was not evaluated in the present study. However, it is important to consider potential advantages and disadvantages of this process vs. the CD process. The longer weld thermal cycles associated with conventional resistance welding would be expected to promote greater conduction of heat into the surrounding matrix, a longer thermal cycle and shallower temperature gradients across the weld zone, thereby promoting greater fiber dissolution and fiber-matrix interactions. However, more precise control over the magnitude and duration of heat input into a conventional spot weld may allow greater control over the solid-state welding process, reduce defects and allow the formation of a softer, more ductile weld zone. Whether improvements in the matrix microstructure resulting from increased process control and slower cooling rates would increase weld strength, despite a greater extent of fiber-matrix interaction, is a subject for future investigation.

Weld Structure/Property/Fracture Relationships

Correlation of the experimental results have clearly shown that if a defect-free solid-state weld is produced between sheets of fiber-reinforced Ti-6Al-4V, then tensile shear strength of the weld is determined by the shear strength of the surrounding HAZ, not the weld interface. In the optimized welds produced between fiber-reinforced sheets (170 V), fracture appeared to initiate at the weld outer periphery notch, and propagate along one or both of the adjacent layers of fibers. The tensile shear fracture stress of these weld specimens averaged 268 MPa (39 ksi), which was about 80% of the fracture stress for a weld produced in the monolithic sheet and which failed in a similar shear mode along or near the weld interface. Interestingly, the dissimilar material welds failed in exactly the same manner at a nearly identical average fracture stress of 270 MPa (39.3 ksi). Based on metallographic analysis, the cross-sectional area of the titanium ligaments between the fibers can be calculated to represent approximately 42% of the total area. Consequently, it is apparent that although fracture along the fiber-matrix interface occurs in relatively brittle appearing manner, the presence of the fibers do contribute to shear strength over that predicted just from the titanium ligaments. A direct influence of the fiber-matrix interface on tensile shear properties of the welds is further indicated by the decrease in shear strength of the welds produced at 180 V. Although these welds exhibited a high-integrity, solid-state welded interface, localized melting around the fibers promoted fracture at an appreciably lower stress level of 199 MPa (28.9 MPa).

The observation of a second fracture mode in the welds produced between the fiber-reinforced sheets at 180 V was of particular interest since the fracture loads and stresses were very comparable to that of the specimen that failed by the more common mode described above. Although fracture initiated at the weld outer periphery, and crack propagation generally occurred along the fiber-matrix interface adjacent to the weld interface as described above (Fig. 128), final fracture occurred through essentially unaffected base metal remote from the weld zone. This result suggests that tensile-shear stress values are determined principally by the initial crack initiation and propagation. Fracture through the unaffected base metal may result from the complex stress distribution during tensile-shear testing of a single-lap specimen, including the presence of appreciable bending stresses.

Conclusions

The following conclusions were developed from present research on the CD resistance spot welding of monolithic and fiber-reinforced Ti-6Al-4V sheet.

1) A systematic parametric study of the CD resistance spot welding process determined that capacitor voltage (and corresponding current) and electrode force could be controlled to reproducibly generate solid-state and fusion spot welds in both monolithic and fiber-reinforced Ti-6Al-4V sheet.

2) Solid-state and fusion welds in Ti-6Al-4V sheet exhibited tensile-shear strengths above minimum requirements set forth in Mil. Spec. MIL-W-6858D. Shear fracture in the solid-state welds occurred through the weld interface region while fracture in the fusion welds occurred by nugget pullout.

3) Solid-state welds were produced in the fiber-reinforced sheet which exhibited complete welding across the weld interface with no evidence of fiber displacement or degradation. Increased voltage promoted the initiation of melting at the fiber/matrix interface and, at sufficiently high voltage, across the entire weld interface. Excessive interface melting promoted appreciable fiber dissolution and displacement.

4) Optimized solid-state welds in the fiber-reinforced material exhibited tensile-shear strength and stress levels 60 and 80%, respectively, of similar welds produced in the monolithic material. This reduction in fracture strength was attributed to fracture along a layer of fibers in the weld HAZ adjacent and parallel to but remote from the weld interface.

5) Dissimilar solid-state and fusion welds were produced between monolithic and fiber-reinforced sheets. The average tensile-shear strength and stress of the welds produced at 160 V were 70 and 80%, respectively, of similar welds produced in the monolithic material. Fracture of all welds occurred remote from the weld interface along an adjacent layer of fibers.

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References


2. Continuous Silicon Carbide Metal-Matrix Composites, Textron Specialty Materials,


