

A fibre coating process for advanced metal-matrix composites

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A fibre coating process has been used to produce continuously reinforced advanced metal-matrix composites with up to 80% volume fraction of SiC fibre. Matrix materials were an α/β titanium alloy (Ti-Al-V), a dispersion-strengthened titanium alloy (Ti-Al-V-Y), a rapid solidification processed aluminium alloy (Al-4.3Cr-0.3Fe), and intermetallic compounds Ti_3Al and TiAl. Thick metal coatings are shown to adhere well to the fibres, no evidence is found for chemical reaction between the coating and the fibre during the coating process, and the coated fibres can be handled and bent without damage. Tensile test data for Ti-Al-V alloy reinforced with 21% SiC fibre show a modulus near to a theoretical prediction, but tensile strength significantly below prediction. Loss of strength is attributed to the formation of a brittle reaction product during hot consolidation. The advantages and potential of the coated-fibre route for MMC production are discussed.

1. Introduction

Continuously reinforced metal-matrix composites (MMCs) may combine the high strength, high stiffness and low density of ceramic fibres with the toughness and resistance to impact damage of metal matrices.

Liquid infiltration of fibre bundles is a well established fabrication method for aluminium alloy MMC. However, this can lead to poor fibre distribution, particularly with large diameter fibres, and various methods have been developed to overcome this problem. In the hot-moulding method [1] fibres are laid-up with alternate layers of metal foil and held in place either with a fugitive organic binder, or woven with a pure metal wire cross-weave and hot consolidated at a temperature between the liquidus and solidus of the matrix alloy. Alternatively, monolayers of fibres are plasma sprayed with an alloy to make a pre-form tape which is then laminated and hot-moulded in a similar way [2].

Cratchley and co-workers [3, 4] in 1964 proposed a method of producing aluminium alloy MMC reinforced with alumina fibres, in which the fibres were pre-coated with aluminium by drawing through a pendant drop of liquid aluminium and then hot-pressed to produce the finished MMC. Similar fibre coating and consolidation processes have been proposed for the production of nickel-based MMCs reinforced with various fibres [5, 6] and aluminium and magnesium alloys reinforced with carbon fibres [7].

Because of the high reactivity of titanium, fabrication methods for titanium-based MMCs are restricted to the solid state. The usual method is that of vacuum-hot-pressing alternate layers of fibres and metal foil [8], at a temperature of approximately 900°C, to produce a fully dense MMC. This fabrication technique is expensive, and has a number of limitations. Various

ceramic fibres have been incorporated in titanium alloys [8], including Al_2O_3 , boron, B_4C -coated boron, B_4C -coated SiC (BORSIC), and SiC. Of these, the principal continuous reinforcement for titanium alloys is now SiC fibre in the size range 100 to 140 μm diameter [8].

This paper describes an electron beam (EB) evaporation technique for coating 100 μm diameter SiC fibre with matrix materials based upon titanium alloys, titanium aluminides, and an aluminium alloy. Because vapour quenching can be considered a rapid solidification process (RSP) [9], all the matrices were rapidly quenched. The Al-Cr-Fe alloy deposited has been produced in bulk form by EB evaporation and condensation and mechanical properties after working have been reported [10].

2. Materials and experimental method

Metal matrices were evaporated by an electron beam and condensed on to continuous 100 μm diameter silicon carbide fibres (Sigma Composite Materials, Tubingen). Property data for the as-received fibre are given in Table I. The experimental arrangement is illustrated in Fig. 1. Eighty fibres were mounted in a cylindrical array on a fibre carrier above a Temescal 10 kW electron-beam heated double evaporation source, and were rotated at 300 r.p.m. One source incorporated a rod-feed mechanism to replenish the evaporation pool. Deposition of matrices on to the fibres took place at a rate of approximately 5 to 10 $\mu m \text{ min}^{-1}$.

TABLE I Data for 100 μm diameter SiC fibre

Density	3.4 g cm^{-3}
Tensile strength	3.5-3.7 GPa
Tensile modulus	420 GPa
Minimum bend radius	4 mm

TABLE II Elemental analysis of matrix coatings

Batch designation	Composition (wt %)					
	Ti	Al	V	Y	Cr	Fe
F1	90.1	5.0	4.9			
F2	90.4	5.0	4.6			
F3	90.5	4.4	4.4			
F4	90.7	3.9	5.0			
F5	91.2	3.0	4.7			
F6	88.1	6.7	2.9	2.3		
F7	82.4	17.4	0.2			
		(27)*				
F8	56.6	41.5	1.9			
		(56)*				
F9		95.4			4.3	0.3

*At %.

At the evaporation rates used in this work the deposits were believed to be reasonably homogeneous, although, some very fine scale ($\lambda < 50$ nm) modulation in the composition of the deposits was expected due to the rotation of the fibre. However, it was anticipated that subsequent hot consolidation would lead to further homogenization. Rotation also produced a slight asymmetry in the coatings, thicknesses being 10% to 15% greater on one side, and this may have affected somewhat the fibre distribution in the finished MMCs.

The compositions of the coatings deposited were determined by electron probe microanalysis (EPMA) and are given in Table II, and the coating process and the coating thicknesses in Table III. Coated fibres F1-F5 were produced using a single evaporation source, rod-fed with Ti-6Al-4V and these deposits had aluminium contents in the range 3.0 to 5.0% and vanadium in the range 4.4 to 5.0%. For F6, yttrium from a separate evaporation source was added to the deposit from a Ti-6Al-4V rod-fed source, to form a dispersion of Y_2O_3 in the titanium alloy by combination with the oxygen in solid solution [11]. Coatings of the two intermetallic compounds Ti_3Al and $TiAl$ (F7 and F8) were made using two evaporation sources, a Ti-6Al-4V rod fed-source and a pure aluminium source: both these deposits therefore contained vanadium. Coatings of the Al-4.3Cr-0.3Fe alloy (F9) were produced by evaporation from a single source, rod-fed with Al-7.5Cr-1.2Fe alloy. Although the small size of the evaporation sources used in the present work did not permit accurate control of deposit

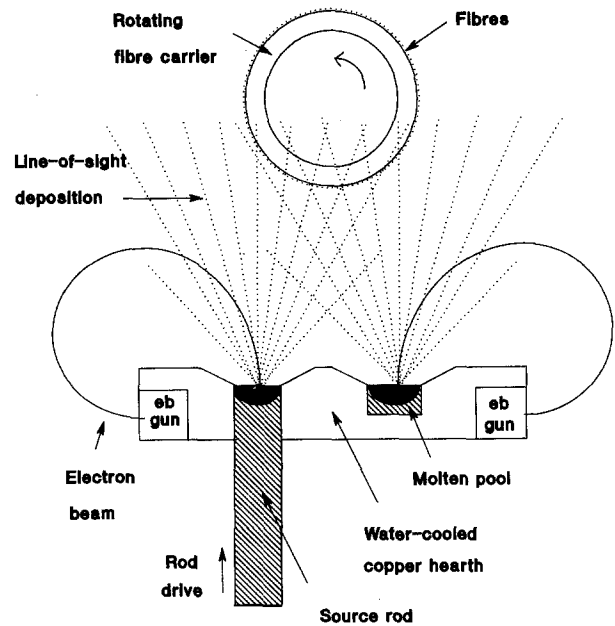


Figure 1 Schematic diagram showing EB matrix coating technique.

composition, under carefully controlled conditions electron beam evaporation can produce deposits of precise and reproducible compositions [10].

During hot consolidation, the coatings deformed and bonded together to form the matrix of the composite. The coating thickness therefore determined the volume fraction of fibre in the finished MMC. In this work, coatings in the range 6 to 65 μ m thickness were deposited on the fibres, producing, on consolidation, MMCs with fibre volume fractions between 19% and 80% (Tables III and IV).

An additional step in the production process was employed for the aluminium-coated fibres, to assist bonding during subsequent consolidation [12]. Immediately after deposition of the matrix coat, and before the coated fibres were exposed to the air, a layer of silver about 1 μ m thick was deposited by EB evaporation. The silver diffused into the aluminium alloy matrix during consolidation.

Consolidation of the coated fibres was carried out either by vacuum-hot-pressing at pressures up to 100 MPa, or by hot-isostatic-pressing (HIPing) at pressures up to 150 MPa. Vacuum-hot-pressing was carried out using a stainless steel lower die containing the coated fibres, and an upper ram which was either

TABLE III Coating process and coated-fibre properties

Batch designation	Nominal composition (wt %)	Evap. source		Coating thickness (μ m)	Minimum bend radius (mm)
		1	2		
F1	Ti-5Al-5V	Ti-6Al-4V	—	35	9
F2	Ti-5Al-5V	Ti-6Al-4V	—	6	9
F3	Ti-4Al-4V	Ti-6Al-4V	—	37	9
F4	Ti-4Al-5V	Ti-6Al-4V	—	25	9
F5	Ti-3Al-5V	Ti-6Al-4V	—	37	9
F6	Ti-7Al-3V-2Y	Ti-6Al-4V	Y	27	12
F7	$Ti_3Al + V$	Ti-6Al-4V	Al	65	22
F8	$TiAl + V$	Ti-6Al-4V	Al	25	32
F9	Al-4.3Cr-0.3Fe	Al-7.5Cr-1.2Fe	Ag*	17	4
Uncoated fibre					4

*Diffusion-bonding interlayer (applied after matrix-coat).

titanium alloy (for production of the titanium-based MMCs), or aluminium alloy (for the aluminium-based MMC). HIPing of coated fibres was carried out in an evacuated and sealed titanium alloy container. A Ti-4Al-4V tensile test-piece (MMC3), longitudinally reinforced with 21% SiC fibre, was produced by vacuum-hot-pressing coated fibres from three batches (F3, F4 and F5) for 2 h at 20 MPa and 925°C. The arrangement for pressing the fibres for the tensile test-piece, and the finished test-piece are illustrated in Fig. 2.

After consolidation, sections normal to the fibre axis were examined by optical and scanning electron microscopy. Matrix hardnesses were determined by Vickers micro-hardness indentation using an 80 g load.

3. Results

3.1. As-coated fibres

In addition to EPMA of each coating (Table II), oxygen analysis carried out on coated fibres taken from batch F3 indicated 6000 p.p.m. oxygen (by weight). However, analysis of batches of uncoated SiC fibre gave 6000, 4700, and 7900 p.p.m. oxygen, indicating a high oxygen level in the fibre, and no conclusion could be made concerning oxygen contamination during coating. Analysis of thick deposits of Ti-6Al-4V alloy, made on a metal substrate at 500°C (other conditions identical to F3), indicated an oxygen pick-up during deposition of about 600 p.p.m. (by weight), which is within acceptable limits for most commercial titanium alloys.

The coated fibres were generally unaffected by handling. For example, all the coated fibres could be wound repeatedly around small radii without any loss of coating or fracture of the fibre. Minimum bend radii for the coated fibres are presented in Table III, together with details of the coating process, and the coating thicknesses.

A scanning electron microscope (SEM) fractograph of a SiC fibre after coating with titanium alloy Ti-5Al-5V, and fractured in bending is shown in Fig. 3. The coating microstructure consisted mainly of columnar grains typical of metals vapour deposited at low fractions of their melting temperatures [13].

However, at higher magnifications, a pore free region was visible adjacent to the fibre surface and extending to a thickness of about 8 µm (A in Fig. 4); the microstructure in this region consisted of fine

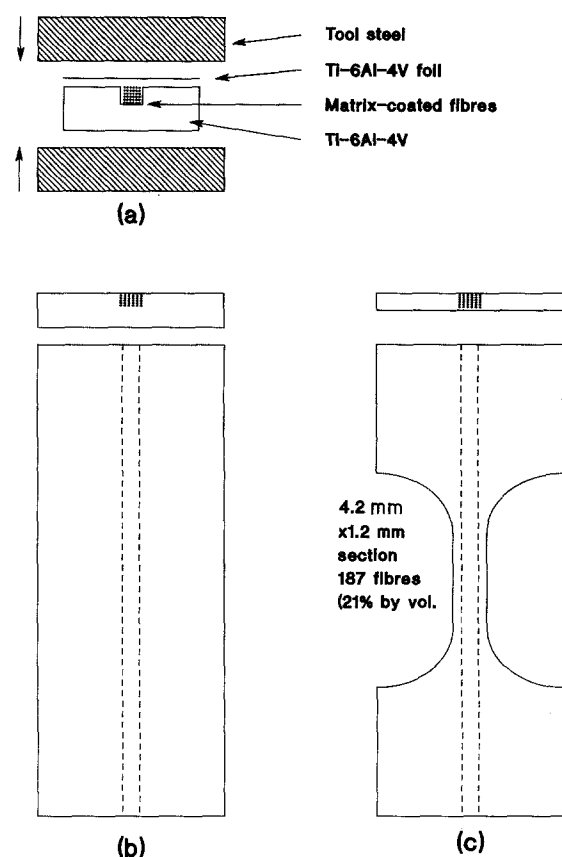


Figure 2 Schematic diagram showing Ti-4Al-4V MMC tensile test-piece manufacture, (a) before and (b) after consolidation; (c) machined test-piece.

equiaxed grains, approximately 1 µm diameter. Above this thickness the coarse columnar grains were orientated with their long axes normal to the axis of the fibre, and the columnar grain boundaries appeared porous in the Ti-6Al-4V-based coatings (F1 to F6), but appeared virtually pore free in the remainder of the coatings (F7 to F9). Part of the fractured end of a Ti-5Al-5V coated fibre is shown in Fig. 5. There was no evidence of chemical reaction occurring between the fibres and the coating during the coating process.

3.2. Consolidated MMC

Consolidation conditions, fibre-volume-fractions, and matrix micro-hardness for the finished MMCs are presented in Table IV. Consolidation experiments were carried out at various pressures; the HIPing pressure given in Table IV is the minimum pressure

TABLE IV Consolidation conditions for coated SiC fibres

Designation	Fibres	Nominal matrix composition (wt %)	Consolidation conditions			Vol. fraction of fibre (%)	Reaction layer thickness (µm)	Matrix microhardness, Hv (kg mm ⁻²)
			Press. (MPa)	Temp. (°C)	Time (h)			
MMC1	F1	Ti-5Al-5V	20	925	2	35	1	369
MMC2	F2	Ti-5Al-5V	150	925	2*	80	1.5	—
MMC3	F3, F4, F5	Ti-4Al-4V	20	925	2	35	1	326
MMC4	F6	Ti-7Al-3V-2Y	150	925	2*	42	1.5	510
MMC5	F7	Ti ₃ Al + V	120	925	2*	19	2	326
MMC6	F8	TiAl + V	150	925	2*	44	1.5	450
MMC7	F9	Al-4.3Cr-0.3Fe	100	500	2	56	0	70
—	—	Ti-6Al-4V	Mill-annealed			—	—	334

*Fibres subject to thermal cycle of 1 h @ 900°C (during canning process, prior to HIPing).

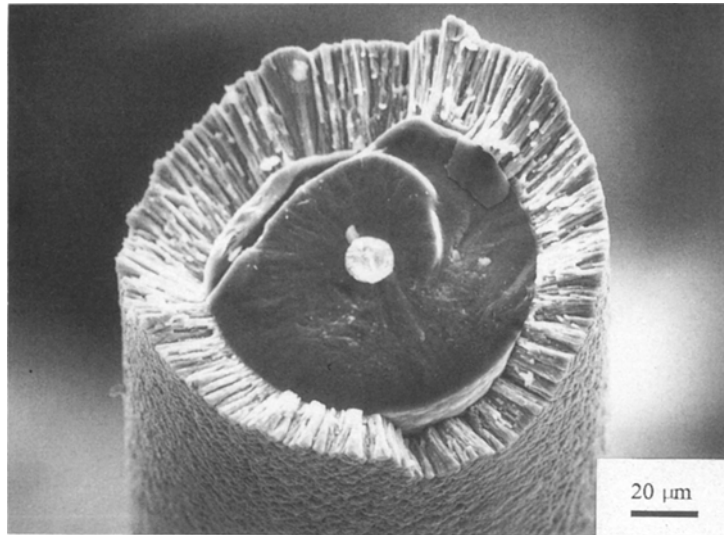


Figure 3 Scanning electron micrograph of SiC fibre, with $\sim 35 \mu\text{m}$ thick Ti-5Al-5V coating (F1).

found for full consolidation (MMCs 2, 4 to 6). However, the optimum consolidation conditions were not determined in the present experiments. The matrix micro-hardnesses of MMC1 and MMC3 (nominally Ti-5Al-5V and Ti-4Al-4V) were near to that of mill-annealed Ti-6Al-4V plate, indicating only slight oxygen contamination; the matrix of MMC5 (Ti_3Al) gave a micro-hardness consistent with its expected strength; the matrices of MMC4 (Ti-7Al-3V-2Y) and MMC6 (TiAl) both gave higher than anticipated micro-hardnesses. The exceptionally high matrix micro-hardness of MMC4 (Ti-6Al-4V-2Y) was attributed to dispersion strengthening by Y_2O_3 precipitates.

The hardness and strength of RSP Al-Cr-Fe alloys increase with increasing chromium and iron content, and decrease with increasing processing temperature. The low hardness value of MMC7 (Al-4.3Cr-0.3Fe) is consistent with the chromium and iron content and the high processing temperature used compared with the values reported elsewhere [10, 14].

Optical micrographs of MMCs with the different matrix materials are shown in Fig. 6. No evidence of either matrix or fibre cracking was found in MMC1 (Ti-5Al-5V), Fig. 6a, MMC3 (Ti-4Al-4V), or MMC5 (Ti_3Al), Fig. 6c. Microstructural inhomogeneity in

the matrix alloy, possibly due to a transient change in deposition conditions during coating, was found in MMC4 (Ti-7Al-3V-2Y), Fig. 6b, together with evidence of cracks in both the fibres and the matrix. Cracking of fibres and matrix was also observed in MMC6 (TiAl), Fig. 6d. In this sample, fine radial cracks in the fibre extending from the fibre surface to a depth of approximately $15 \mu\text{m}$ suggested a proportion of the fibres had suffered from some form of chemical attack. The aluminium alloy fibres (F9) were successfully diffusion bonded to produce a pore-free composite (MMC7), Fig. 6e, with no matrix cracking, and no evidence of the silver layer; some fibre cracking was, however, apparent.

Approximate, fibre/matrix reaction layer thicknesses were measured optically, and these results are presented in Table IV. For titanium alloys the reaction layer thickness were in the range 1 to $2 \mu\text{m}$.

A scanning electron micrograph of an MMC with 80% volume fraction of SiC in Ti-5Al-5V (MMC2) is shown in Fig. 7. This demonstrates the very high volume fraction of fibre and excellent fibre distribution possible with the matrix-coating process.

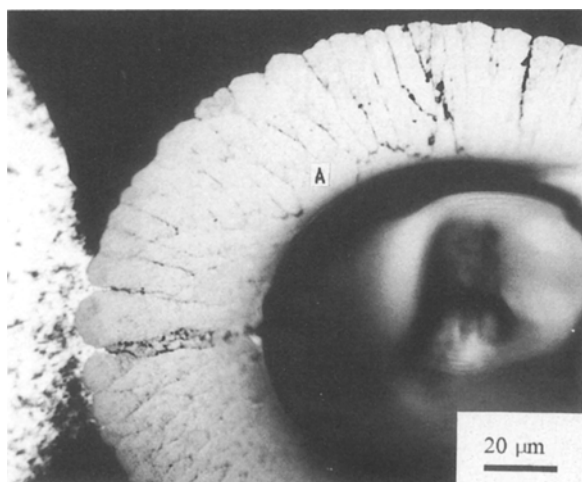


Figure 4 Optical micrograph of a section through SiC fibre coated with Ti-5Al-5V (F1).

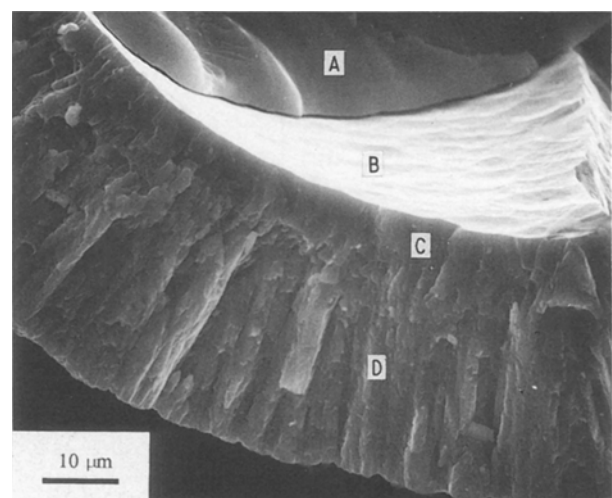


Figure 5 Scanning electron micrograph of a fractured Ti-5Al-5V coated fibre (F1), showing end of fractured SiC (A), inner surface of coating (B), fine non-porous region of coating (C), and columnar growth region (D).

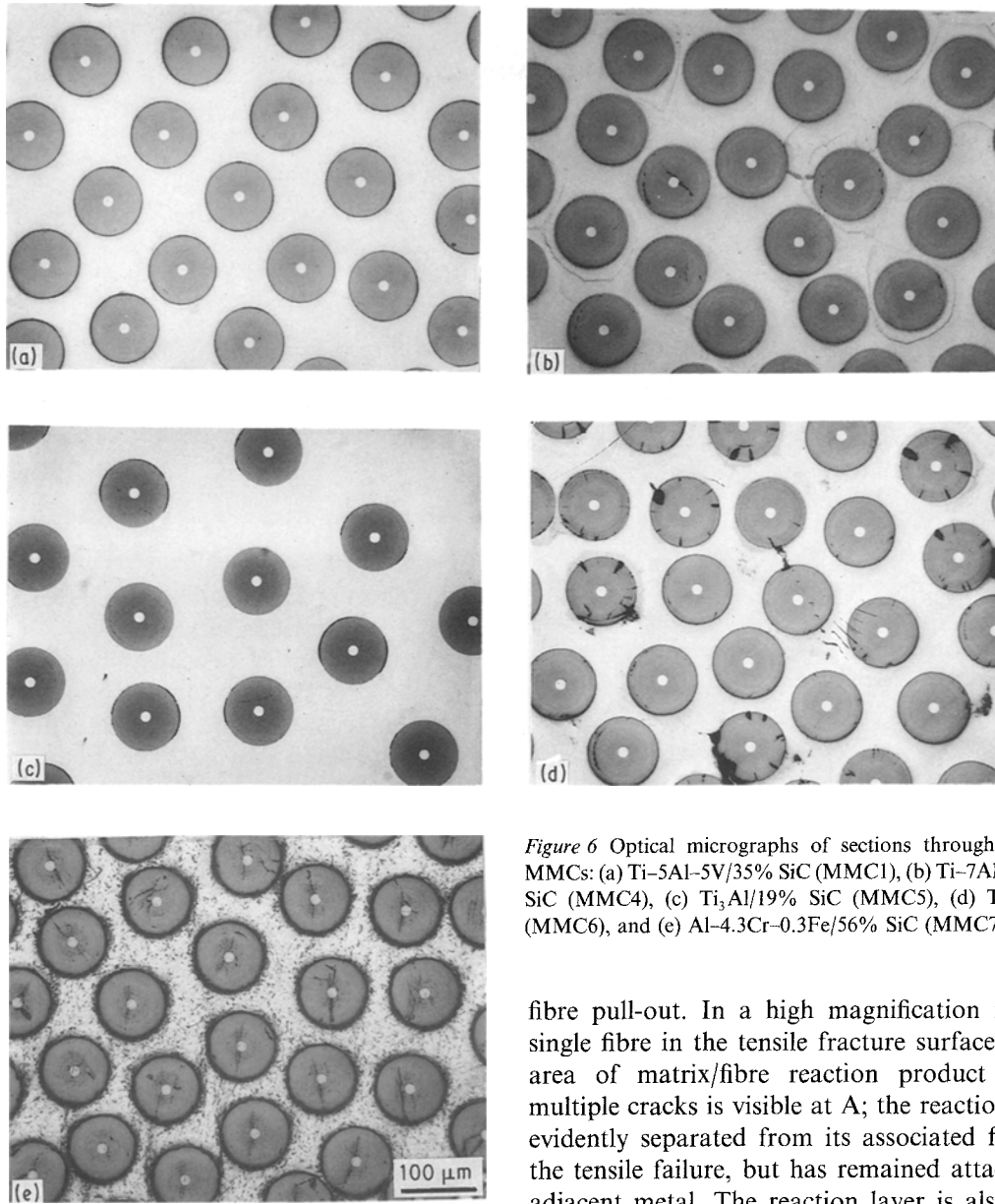


Figure 6 Optical micrographs of sections through consolidated MMCs: (a) Ti-5Al-5V/35% SiC (MMC1), (b) Ti-7Al-3V-2Y/42% SiC (MMC4), (c) Ti₃Al/19% SiC (MMC5), (d) TiAl/44% SiC (MMC6), and (e) Al-4.3Cr-0.3Fe/56% SiC (MMC7).

Tensile test results for Ti-4Al-4V alloy reinforced with 21% SiC fibre (MMC3) are presented in Table V, together with predicted tensile properties, based on the rule of mixtures [15] and using manufacturers data for Ti-6Al-4V alloy and SiC fibre. Young's modulus was near to the calculated value for Ti-6Al-4V/21% SiC, whereas tensile strength was only 54% of the calculated value. The tensile fracture surface is shown in Fig. 8, the surface is relatively planar, with little

TABLE V Tensile test results for MMC3 (Ti-4Al-4V/21% SiC), and predicted tensile properties for Ti-6Al-4V/21% SiC MMC

	Tensile test data	Theoretical prediction*
Young's modulus, E_c (GPa)	153	176
Failure stress, σ_c (MPa)	805	1484

* Calculated according to rule of mixtures (equal strain) [15].

$$\text{Young's modulus } E_c = E_f(V_f) + E_m(1 - V_f)$$

$$\text{Failure stress } \sigma_c = \sigma_f(V_f) + \sigma_m(1 - V_f)$$

where E_f is the fibre modulus (420 GPa), E_m the matrix modulus (110 GPa), V_f the fibre volume fraction, σ_f the fibre tensile strength (3500 MPa), and σ_m the matrix stress at failure strain of fibre (917 MPa).

fibre pull-out. In a high magnification image of a single fibre in the tensile fracture surface, Fig. 9, an area of matrix/fibre reaction product containing multiple cracks is visible at A; the reaction layer was evidently separated from its associated fibre during the tensile failure, but has remained attached to the adjacent metal. The reaction layer is also visible in Fig. 10, and appears to be approximately 2 μm thick. A radial crack, shown at B in Fig. 10, coincides with a crack in the fibre, but has not penetrated the matrix.

4. Discussion

4.1. Comparison of MMC production methods

Matrix coating and consolidation has been demonstrated for the production of silicon carbide-reinforced titanium alloy and aluminium alloy MMCs. This process has a number of advantages over the established MMC production methods.

The present process avoids the use of titanium alloy foil, which is expensive and for many alloys unobtainable, and a much better fibre distribution is possible compared with alternative processes. A recent report on the examination of a Ti-15V-3Cr-3Al-3Sn (usually called 15.3) alloy/36% SiC composite, produced by the foil lay-up process [16], states that 6% of the fibres were touching, and where they touched, the fibres were degraded by micro-cracking. Furthermore, microstructural inhomogeneity was introduced by the pure titanium wire used in the fibre weave.

Consolidation of matrix-coated fibres is likely to be

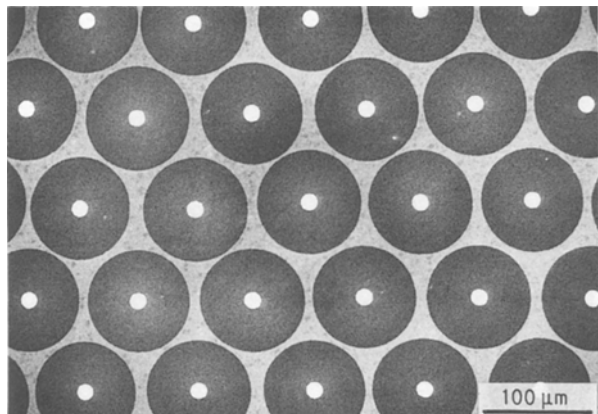


Figure 7 Scanning electron micrograph of section through Ti-5Al-5V/80% SiC composite, after consolidation (MMC2).

easier than consolidation of alternate layers of foil and fibre. Because each fibre is already surrounded with matrix material, the requirements for metal flow will be less severe, and the fine grain microstructure obtained by rapid quenching in the present experiments is known to favour superplasticity and enhance diffusion bonding [17]. Therefore, it may be possible to use lower consolidation temperatures and pressures, or shorter times, with a commensurate reduction in fibre/matrix chemical reaction, or in the number of damaged or broken fibres.

Plasma spraying has been used to introduce the matrix material into fibre-reinforced MMCs [2], and recently this has been demonstrated for titanium alloys by Kieschke and Clyne [18]. However, compared to the matrix-coating process, plasma spraying is likely to be difficult to control and would probably be limited to low volume fractions of fibre. There is concern also about the possibility of gas entrapment during plasma spraying, and the powder or wire required by a plasma gun are both difficult to obtain for many potential matrix materials.

The matrix-coating process gives easy control over fibre volume fraction in the finished MMC, and is capable of much higher fibre volume fractions than alternative methods. The practical limit for the foil-lay-up method of producing titanium-based MMCs is thought to be about 50% volume fraction of fibre, and is in practice about 35%. This study has demonstrated

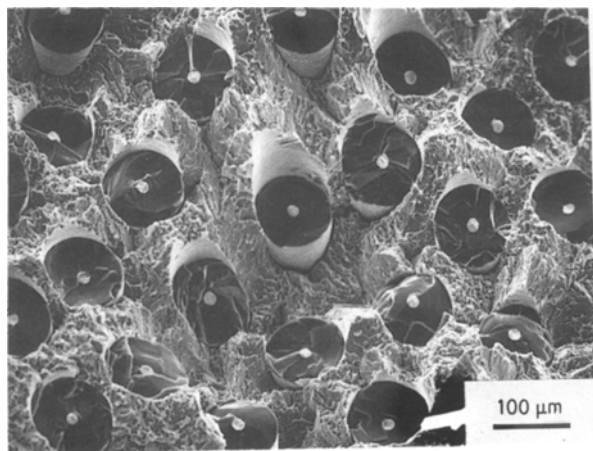


Figure 8 Scanning electron micrograph of Ti-4Al-4V/21% SiC tensile fracture (MMC3).

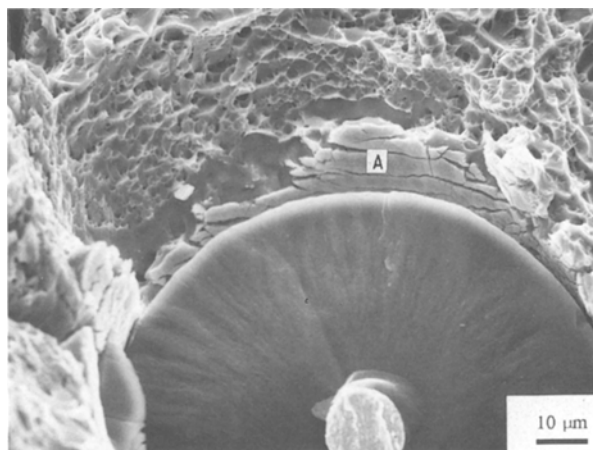


Figure 9 Scanning electron micrograph of Ti-4Al-4V/21% SiC tensile fracture (MMC3), showing fibre/matrix reaction product (A).

the production of titanium-based MMCs with up to 80% volume fraction of fibres.

Consolidation of pre-coated fibres offers an easier and more versatile method of producing complex shapes or contours than the alternative production methods. Coated fibres have been found to be capable of being handled and formed round small radii (see Table III) without damage. Therefore, the use of suitably shaped dies would enable complex near-net-shape MMC components to be produced in a single operation, and the use of well-established filament winding technology, together with HIPing, would facilitate the production of reinforced tubular structures.

A major advantage of the matrix-coating process is that it can be applied to fibres of any diameter, whereas the foil-lay-up process may prove to be prohibitively expensive for fine fibres. It is recognized that for certain MMC applications a fibre diameter of 100 μm may be too large, and it is expected that smaller diameter single-strand silicon carbide fibres will be available in the future (only two types of SiC monofilament have been available to date; the 100 μm diameter Sigma fibre used in this study, and 140 μm diameter fibre produced by Textron Specialty Materials).

In addition to applying a matrix coating, the PVD coating process used in this study could be used to

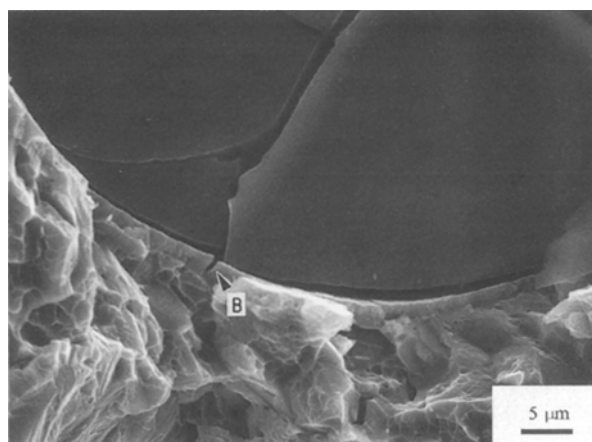


Figure 10 Scanning electron micrograph of Ti-4Al-4V/21% tensile fracture (MMC3), showing fibre/matrix reaction layer, with crack at B.

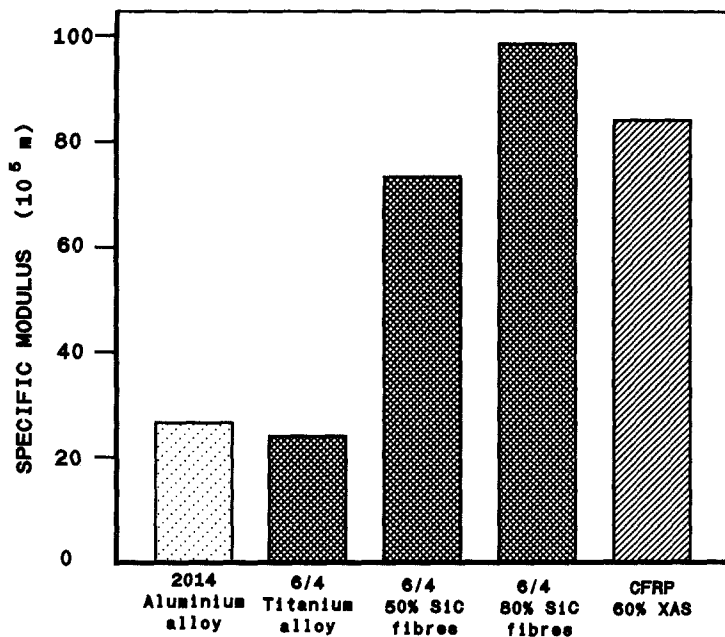


Figure 11 Calculated longitudinal specific moduli for continuously reinforced Ti-6Al-4V, compared with a range of light-weight materials.

deposit an initial barrier layer coating, to reduce chemical reaction between fibre and matrix. Although some reaction is required to give adequate adhesion, it is known that thick reaction layers can cause premature failure to composites [8]. This probably arises because these layers contain defects which give rise to cracks, which propagate into the fibre. For the titanium alloy/SiC composites the reaction products are likely to be Ti_5Si_3 and TiC , which could be strong in thin layers, but brittle in thicker layers, especially in the presence of cracks.

Alternative methods of reducing chemical reaction between fibre and matrix are to lower the forming temperature, either by increasing the forming pressure/time, or by using alloys with lower superplastic forming temperatures [19]. In addition, modifications to surface composition during fibre manufacture have been carried out for the fibre Textron SCS6, which has a carbon-rich surface layer, and reported reaction layer thicknesses of only 0.10 to 0.35 μm for a titanium alloy/SCS6 composite.

4.2. Tensile properties

It has been shown that the matrix-coating method is capable of producing composites with up to 80% volume fraction of fibre. A titanium alloy MMC with this proportion of fibre would be expected to have a specific stiffness superior to that of unidirectional carbon fibre-reinforced plastic (CFRP). This is illustrated by the bar-chart shown in Fig. 11, where the specific stiffnesses of a range of light-weight materials are compared.

The tensile test data for the test-piece produced by the matrix-coated fibre process showed a Young's modulus close to the predicted value, but tensile strength considerably lower than the calculated value. This discrepancy is attributed partly to the fact that the matrix of MMC3 was deficient in aluminium compared with Ti-6Al-4V, the alloy used for the prediction calculation, but mainly to excessive chemical reaction between the fibre and the matrix during consolidation at elevated temperature.

4.3. Potential matrix materials

The matrix-coating process can be applied to any alloy capable of being deposited by evaporation, and which is amenable to consolidation by hot pressing, e.g. aluminium-, titanium- and nickel-based composites. This can include conventional alloys, or alloys which are impossible to produce by orthodox ingot metallurgy, such as titanium alloys containing rare-earth or metalloid additions [20-24].

The vapour-deposition process has the advantage of having a very high effective solidification rate. Rapid solidification processing (RSP) offers a possible route for the improvement of titanium alloy mechanical properties, both by refinement of the α/β microstructure, and by the introduction of fine dispersed phases. Effort to date on the development of RSP titanium alloys has concentrated on melt spinning, and atomizing [20-24], but, results have been variable, mainly due to inhomogeneous microstructures.

Elements under consideration as dispersion formers in titanium alloys include rare earth elements yttrium, cerium, dysprosium, erbium, gadolinium, lanthanum or neodymium, which on ageing form dispersoids in titanium by internal oxidation, and the metalloid elements boron, carbon and silicon, which form TiB , TiC , and Ti_5Si_3 . In the as-quenched condition, titanium alloys containing these additions show extended solid solubilities, and grain refinement, with up to 47% increase in hardness being reported [22]. Ageing produces fine second-phase dispersions, and in most cases a significant hardening response is observed; the degree of hardening depending upon the system. Work has shown that for additions up to about 4% (atomic) the rare earth elements are much more effective age hardeners per atomic per cent than the metalloid elements [22].

Quench rates for vapour deposition have been estimated to be $10^{13} \text{ K sec}^{-1}$ [9], compared with 10^4 to 10^8 K sec^{-1} for liquid-quenching methods. Therefore, the potential exists with vapour quenching for producing alloys with greatly extended solid solubilities, or very fine dispersions of second phases.

The example of a dispersion-strengthened alloy chosen for this study was Ti-7Al-3V-2Y. In aged RSP titanium alloys, yttrium forms a fine dispersion of the oxide Y_2O_3 , and this has been shown to have excellent stability in α -titanium [11]. The yttrium-strengthened matrix of MMC2 gave a micro-hardness considerably higher than both mill-annealed Ti-6Al-4V and matrix-coated and consolidated Ti-5Al-5V (MMC1).

In this study the titanium aluminides Ti_3Al (MMC3) and TiAl (MMC4) have been coated on SiC fibres and consolidated into MMC. Intermetallic compounds are attractive both as bulk materials, and as matrices for fibre-reinforced composites. The two titanium aluminides, Ti_3Al and TiAl, have very high moduli, reasonable strengths, low densities relative to titanium and nickel alloys, and the ability to retain their properties at elevated temperatures [25-27]. In addition, they both have good oxidation resistance compared to titanium alloys. The principal difficulty in the application of these aluminides is their relatively poor tensile ductilities [25-27].

Various alloy additions have been made to improve the ductility of titanium aluminides, notably niobium to Ti_3Al [28] and vanadium [29] or boron [30] to TiAl. Niobium is difficult to deposit by evaporation, due to its relatively low vapour pressure. However, the RSP vapour-quenching route used in this study to coat the fibres has the potential to improve the ductility of intermetallic compounds by grain refinement, and the inclusion of fine dispersoids such as rare earth oxides [31].

4.4. Potential mechanical properties

The properties of high-strength ceramic fibres have been exploited successfully in the form of fibre-reinforced plastics. However, the use of a metal matrix brings a number of major advantages; these include high-temperature capability, abrasion and damage resistance, environmental stability and significant transverse strength.

Titanium alloys exhibit a rather low specific stiffness compared with aluminium alloys and steels. However, using a rule of mixtures calculation, reinforcement with 35% SiC fibre has the potential to more than double the longitudinal specific modulus of Ti-6Al-4V, and similarly, specific strength would be expected to increase by slightly less than a factor of two.

The two titanium aluminides TiAl and Ti_3Al both have specific moduli almost twice that of Ti-6Al-4V, but have relatively poor specific tensile strengths. Basic physical and mechanical properties of the two titanium aluminides are presented in Table VI, together with comparative data for Ti-6Al-4V. Calculation suggests that reinforcement with 35% SiC would increase

TABLE VI Properties of potential matrix materials

	Density, ρ ($g\ cm^{-3}$)	Fracture stress, σ_f (MPa)	Young's modulus, E (GPa)
Ti-6Al-4V	4.42	900	110
Ti_3Al	3.06	537	147
TiAl	3.80	472	171

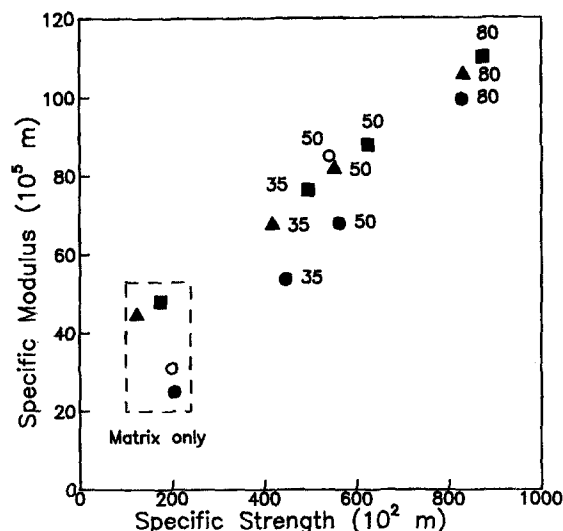


Figure 12 Calculated specific moduli plotted against specific strengths for (●) Ti-6Al-4V alloy, (■) Ti_3Al and (▲) TiAl intermetallic compounds, and (○) Al-Li(8090) alloy, in bulk and reinforced with different volume fractions of SiC fibre.

specific strengths of TiAl and Ti_3Al to near that of a Ti-6Al-4V/35% SiC composite, and would increase specific moduli to ~40% above that of Ti-6Al-4V/35% SiC. A plot of calculated specific strength against specific modulus for composites with Ti-6Al-4V, titanium aluminide and aluminium-lithium alloy matrices, and various volume fractions of SiC fibres is presented in Fig. 12 (all composite properties calculated according to rule of mixtures).

Aluminium alloys are not normally used above 200°C and titanium alloy Ti-6Al-4V is not normally usable above about 450°C. Aluminides Ti_3Al and TiAl retain over 85% of their room-temperature strength and stiffness at 800°C [25-27]. The effect of temperature on the calculated specific moduli and specific strengths of Ti-6Al-4V, Ti_3Al and TiAl bulk materials, and 35% SiC-reinforced composites are presented in Figs 13 and 14 (also shown for comparison in Fig. 14, are data for the RSP Al-Fe-V-Si alloy FVS1212 [32], which shows exceptionally good elevated temperature properties for an aluminium

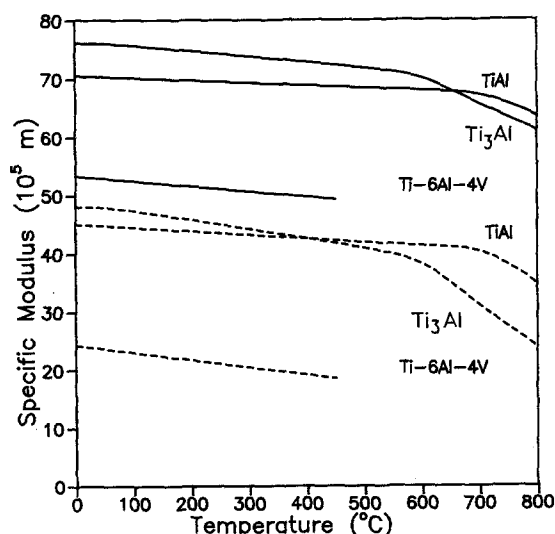


Figure 13 Specific modulus plotted against temperature for Ti-6Al-4V alloy, and Ti_3Al and TiAl intermetallic compounds, in (---) bulk and (—) reinforced with 35% SiC fibre.

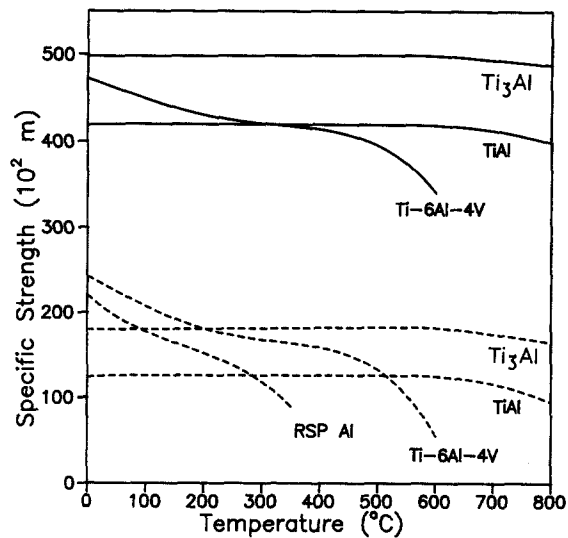


Figure 14 Specific strength plotted against temperature for Ti-6Al-4V alloy, and Ti_3Al and $TiAl$ intermetallic compounds, in (---) bulk and (—) reinforced with 35% SiC fibre (RSP aluminium alloy FVS1212 also shown).

alloy, but is limited to about 350°C). A titanium-based composite may be degraded by long-term exposure above about 600°C.

5. Conclusions

1. Electron-beam evaporation may be used to coat ceramic fibres with matrix alloy. During hot consolidation, the coating deforms and bonds to form the matrix of the composite.

2. No chemical reaction is observed to occur between SiC fibre and titanium alloy during coating; alloy coatings up to 65 μm thickness adhere well to SiC fibre, and coated fibres may be handled and bent without damage.

3. Fibres coated with a Ti-Al-V α/β titanium alloy, intermetallic compound Ti_3Al , or RSP aluminium alloy Al-4.3Cr-0.3Fe were consolidated into fully dense composites, with no evidence of matrix cracking. MMCs produced from fibres coated with dispersion-strengthened alloy Ti-7Al-3V-2Y, or TiAl, showed some matrix cracking.

4. It has been found that, after consolidation, compared with other methods, the matrix-coated-fibre process produces MMCs with a good fibre distribution, and no fibre/fibre contact.

5. Composites with 80% volume fraction of fibre can be produced by the matrix-coated fibre process.

6. The Young's modulus of a titanium alloy MMC with 21% SiC was found to be near to the theoretical prediction, but tensile strength was significantly less than predicted. Loss of strength was attributed to the formation of a brittle reaction layer between the fibre and the matrix during hot consolidation.

Acknowledgement

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