

Flexure and tension tests of newly developed ceramic woven fabric/ceramic matrix composites

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새로 개발된 세라믹 직포 보강 세라믹 기지 복합체의 인장 및 곡강도 시험

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Abstract The mechanical properties of 2D ceramic composites fabricated by the newly developed powder infiltration and subsequent multiple impregnation process were characterised by both 3-point flexure and tensile tests. These tests were performed with strain gauge and acoustic emission instrument. The woven fabric composites used for the test have the basic combinations of Al_2O_3 fabric/ Al_2O_3 and SiC fabric (Tyranno)/SiC. Uniaxially aligned SiC fibre (Textron SCS-6)/SiC composites were also tested for comparison. The ultimate flexural strength and first-matrix cracking stress of SiC fabric/SiC composite with 73 % of theoretical density were about 300 MPa and 77 MPa respectively. However, the ultimate tensile strengths of composite were generally one third of flexural strengths, and first-matrix cracking stress in a tension test was also much lower than the value obtained from flexure test. The lower mechanical properties measured by tension test were analysed quantitatively by the differences in stressed volume using Weibull statistics. This showed that the ultimate strength and the first-

matrix cracking stress of woven laminate composites were mainly determined by the gauge length of fibres and the stressed volume of matrix respectively. Incorporation of SiC whiskers into the matrix increased first-matrix cracking stress by increasing the matrix failure strain of composites.

요 약 새로 개발된 분말침투 및 연속 다중침투법에 의해 제조된 세라믹 섬유 복합체의 기계적 물성을 3점 곡강도 및 인장 시험을 통하여 평가하였다. 정확한 물성 측정을 위하여 strain gauge 및 acoustic emission 측정 장비가 사용되었다. 실험 시편은 Al_2O_3 직포/ Al_2O_3 와 SiC 직포/SiC를 기본 재료로 하고 있으며, 일방향으로 배열된 SiC 섬유(Textron SCS-6)/SiC 복합체를 비교 목적으로 제작 시험하였다. 이론 밀도의 약 73%인 SiC 직포/SiC 복합체의 최대곡강도는 300 MPa이고, 기지내 균열이 처음 발생하는 응력은 77 MPa였다. 인장강도는 곡강도의 1/3 정도의 낮은 값을 나타내었고, 인장 시험중의 첫번째 기지 균열 응력 또한 곡강도 시험에서 얻은 값보다는 상당히 낮은 값을 보여주었다. 곡강도 물성에 비교하여 상대적으로 낮은 인장물성은 Weibull 통계 처리 방법에 의하여 응력을 받고 있는 부피의 차로 정량적으로 해석하였다. 해석 결과, 직포가 층으로 배열된 복합체의 최대 인장강도는 응력을 받는 섬유의 길이에 의존하며, 기지내 균열이 생기는 첫번째 응력은 응력을 받는 부피에 의해 결정됨을 보여주었다. SiC 휘스커를 기지에 보강함으로써 복합체의 기지파괴 strain을 향상시키는 이유로, 첫번째 기지 균열 응력이 증가됨을 확인하였다.

1. Introduction

The flexural test has been used most extensively for evaluating the mechanical properties of fibre-reinforced ceramic composites because of its simple specimen geometry, easy alignment, and relatively small specimen required for standard testing. In an elastic beam with rectangular cross-section, under a bending load, there is a linear variation of longitudinal stress and strain from the compressive side to the tensile side, through zero at its neutral plane. The maximum stress at failure occurring at those surfaces can be computed from the bending moment, and is known as the modulus of rup-

ture (MOR). When a rectangular bar is loaded in 3-point bending the maximum tensile (σ_t) and compressive (σ_c) stresses occur at the surfaces under the middle loading point. The maximum shear stresses (σ_s) occur on the neutral plane.

Therefore the mode of failure depends on the relative value of tensile, compressive and shear strengths. Thus if approximate values of these strengths are known then an appropriate ratio of span to thickness can be defined to introduce the assumed failure mode [1]. Marshall and Evans [2] observed that failure occurred at the compressive side of the beam by buckling of fibres and fragmentation of the matrix in 4-point flexural test

when the separation of the inner and outer loading points to the beam thickness is larger than 8. However, when the ratio was smaller than 5 for the same materials, failure occurred by cracking along the neutral plane between the inner and outer loading points where shear stresses are highest. This change of fracture mode is consistent with the change in the ratio of the maximum applied compressive (or tensile) stress to the maximum applied shear stress as loading configuration changes. This observation could indicate at least two points. First, unlike a monolithic ceramic, the compression strength of a ceramic matrix fibre composite may sometimes be similar to, or even less than the tensile strength. Secondly, the shear strength of planes parallel to fibres are generally less than the tensile or compressive strength parallel to fibres.

The failure stress obtained from flexure test is in general greater than tensile strength because of existence of flaws and the greater probability of finding a critical flaw in the larger stressed volume of a tensile specimen. Moreover, if there is matrix cracking in the tensile side of bending beam the difference is enhanced because, although the strain still varies linearly across a section, the stress in the cracked part of tensile zone will be approximately constant and equal to the matrix cracking stress. As a result the neutral plane of the beam will move towards its compressive face, giving an increase in bending moment and a high MOR. Aveston et al. [3] considered the matrix cracking effect on the flexural strength

of the fibre-reinforced ceramic composites, and showed a way to derive the true value from apparent MOR for a brittle matrix composite. The relationship between the ratio (R) of apparent and true MOR is a complex function of parameter $\alpha = E_m V_m / E_f V_f$, whose form depends on the ratio of ultimate strains of fibre and matrix ($\epsilon_{fu} / \epsilon_{mu}$). In the case of Nicalon fibre and pyrex matrix composite, $\epsilon_{fu} / \epsilon_{mu}$ is ~ 13.5 because $\epsilon_{fu} = 1.35\%$ and $\epsilon_{mu} = 0.1\%$, and $\alpha = 0.6$ gives R of 1.1. Thus, the true MOR can be about 10% less than apparent MOR.

Therefore, a flexural test could give rise to a significantly higher ultimate failure stress in the fibre-reinforced ceramic matrix composites in comparison with the value obtained from pure tension test. As a consequence, stress levels on the tensile side can no longer be derived by assuming simple beam theory. The use of strain-gauge analysis to derive the separate stress-strain behaviour in tension and compression can in principle cope with shifts of the neutral axis, and thus is potentially a valuable tool. It can certainly illustrate well the onset of matrix cracking, but its reliability can be questioned because of the practical limitations of the effects of random cracking under the limited area of each gauge.

Although the fibre-reinforced ceramic composites tested under flexural loading can fail other than in tension and exhibit overestimated MOR, these problems can be relieved by increasing the span-to-thickness ratio, but require careful interpretation of load-deflection behaviour beyond linearity

in the curve. Obviously, the load-deflection curve beyond the initial elastic regime in flexural test must be considered as only 'nominal' values because of the complications arising from matrix cracking on stress distribution in the three point flexure geometry.

Uniaxial tensile testing of fibre-reinforced ceramic matrix composites could alleviate most of the problems associated with flexural tests. Therefore, for acquisition of design data, the tensile stress-strain behaviour is considered the most important single property for ceramic matrix based composites. Under uniaxial tensile load, the material experiences uniform tensile stress in the tested volume and it fractures when its ultimate tensile strength is reached, and then the tensile stress can be obtained from the measured load divided by the cross-sectional area of the test section. However, considerable care must be taken to ensure that the test is meaningful. The main problems in measuring the tensile strength of a fibre composites are obtaining good alignment and ensuring that a valid tensile failure occurs without premature splitting parallel to the fibres. Splitting can occur either because of shear stresses arising from the specimen geometry or, if compressive wedge grips are used, by indentation of the grip surface serration into the specimen. In order to avoid these problems, tensile testing requires unique gripping and specimen design, but at present the difficulties associated with testing reliability have precluded standardisation of any tensile methods [4].

Tensile testing has been performed for the SiC fibre/pyrex composite [5-7] and carbon fibre/pyrex glass composites [7,9]. All of tensile testing for these composites were carried out using parallel-sided flat specimens. None of these mentioned the observation of fracture-initiation site and the verification of alignment, except for the tests performed by Morrell and McCartney [9]. They described that in the majority of cases, one part of the fracture extended to just under the tabbed area, but whether it originated at this point was unresolved. This report indicates that the gripping and axiality problems should be attenuated by the most effective test-piece design.

From the above discussion, neither a tension test with small parallel-sided specimen nor a flexure test can present adequately characterise the properties of ceramic matrix fibre composites. A valid tensile test inevitably requires the development of test-piece design, and verification of alignment on an adequate and standardisable basis. The objective of this study is to introduce the new technique for tensile test of the newly developed 2D ceramic fibre composites.

2. Experimental procedure

Materials used for this study are shown in Table 1. SiC or Al₂O₃ powders mixed with 5 to 10 wt % of as-received PCS (polycarbosilane), respectively, were ball milled with ethanol using a centrifugal jar for 20

Table 1

Raw materials

	Materials	Manufacturing company	Characteristics
Binder	Polycarbosilane	Shinetou Chemical Co., Japan	Mn ⁽¹⁾ ; 1400-2000, M.P.; 210°C - 250°C, After heating up to 1200°C in N ₂ gas; weight loss: 40-45 wt%, chemical compositions: Si 60 w/o, C 40 w/o and 0 (<1w/o), density: 3.2 g/cm ³
Matrix	Al ₂ O ₃	Alcoa, USA.	SG-16
	SiC	Mitsui Toatsu Chemical Co., Japan	Ave. particle size; 0.15 μm, Impurity; ~0.02 w/o, Density; 3.21 g/cm ³
	SiC whisker	American Matrix Inc., USA	β-SiC, Dia.; 1-3 μm, Length; 20-300 μm, Density; 3.21 g/cm ³
Fabrics	SiC(Tyranno)	Ube Industries Japan	Plane fabric, 1600 fibres/yarn, Fibre; density: 2.3-2.4 g/m ³ , dia.: 8.5 μm, Weave density; warp: 15 yarns/inch, weft: 15 yarns/inch, weight: 260 g/m ² .
	Al ₂ O ₃ (Almax)	Mitsui Mining Co., Ltd., Japan	Plane fabric, 1000 fibres/yarn, Fibre; density: 3.6 g/m ³ , 99.5% of α-Al ₂ O ₃ , dia.: 10 μm, Weave density; warp: 15 yarns/inch, weft: 15 yarns/inch, weight: 300 g/m ² .

(1) A number of average molecular weight.

hrs, which resulted in viscous slurries. In case of SiCw containing matrix, the matrix mixture without SiCw was first milled for 20 hrs, and then further mixing was carried out with 20 wt% of SiCw for 2 hrs using teflon balls, in order to reduce whisker damage during mixing. Woven fabrics cut to 50 × 50 mm size were dipped into matrix slurry, and then the slurry containing fabric sheets were stacked 4 or 5 layers into a porous container. Infiltration of matrix slurry between fibre and fibre in a yarn was en-

hanced by applying about 5 to 10 MPa pressure to the stacked laminate sheets.

Dried composite compact was consolidated by induction heating in nitrogen atmosphere. The graphite cylinder with 5 cm × 5 cm rectangular hole was used for mould pieces. The processing conditions, e.g. temperature, time, pressure, heating and cooling rates, and atmosphere are optimised to produce a dense composite with little matrix cracking. Details of optimised heat treating procedure has been described in elsewhere [10].

For convenience, the first heat treated specimen is called 'as-fabricated', and the samples obtained from further impregnation and heating procedure are named by number of impregnation and heating procedures, e.g 'one time impregnated' or 'two times impregnated' etc. As-fabricated composite tiles were cut into samples with dimension of 4 mm width, 40 mm length and ~ 2 mm thickness using diamond saw.

The theoretical density of composite is also calculated with assumption that all the pores in the as-fabricated sample are filled with product of pyrolysed product of PCS, i.e, β -SiC. In case of as-fabricated Al_2O_3 fabric/SiCw + Al_2O_3 composite with weight of 10 g, fibre loading of 16.7 vol% can contribute 3 g, and then remaining 7 g is due to the weight of Al_2O_3 , SiCw, and β -SiC resulted from pyrolysis of PCS. In the original powder mixture, 20 wt% of SiCw to Al_2O_3 powder was mixed, and if 10 wt% of PCS was mixed to SiCw and Al_2O_3 mixture, then the weight of pyrolysed product is 0.434 g from $0.1 (10 \text{ wt}\%) \times 0.62$ (weight residue 62 %) $\times 7$ (weight of matrix). The remaining 6.566 g of matrix consists of 1.313 g of SiCw and 5.253 g of Al_2O_3 . The volume fraction of pyrolysed product of PSC, Al_2O_3 and SiCw are 3.3, 31.6 and 9.8 %, respectively. Taken 16.7 vol% of fibre and assuming 38.6 vol% of pores are filled with β -SiC, the theoretical density of this composite is 3.30 g/cm^3 . The calculation of theoretical density of SiC fabric/SiCw + SiC composite is rather straight forward, since all matrix constituents have same density of 3.21 g/cm^3 .

Three-point bend test was performed in 30 mm span with cross head speed of 0.5 mm/min at room temperature. The load was applied perpendicular to the layers of SiC cloth. The span to depth ratio is about 15 for the composite specimens. Ultimate flexural strength and proportional limit were obtained from peak stress, and initial linear portion of load-deflection curve, respectively. In order to identify first-matrix cracking stress at which matrix cracking begins, some of woven laminate composites and monolithic ceramics were monitored with acoustic emission instrumentation during flexural test. Strain gauge with 2 mm gauge length was also attached in the middle of tensile surface where fracture was always initiated. The acoustic emission together with applied stress level were recorded as function of time by disitising oscilloscope. The stress-strain curve was separately recorded. First matrix cracking strain of composite was obtained from the strain level at which the major acoustic emission was first identified. Elastic modulus of composite was determined by initial linear portion of stress-strain curve.

The data obtained from flexure test cannot be used for design purpose, because the results of this test cannot be interpreted in a consistent manner due to the potential presence of compressive buckling and shear failure. However, tensile test requires specific gripping and specimen design to provide good alignment and a valid tensile failure at a necked gauge-length. Compressive wedge grips can cause premature splitting, and

even small misalignment can give rise to a significant torsion and bending stress during testing. Therefore, unique fixture and specimen configuration were developed to provide meaningful tensile data.

The fixture configuration which could avoid any specimen damage due to compressive gripping stress are shown in Fig. 1. The assemble of tensile specimen with fixture is demonstrated in Fig. 2

In order to verify alignment, strain gauges were attached to each surface of necked zone, thus 4 strain gauges for each specimen were used. Then, the test was carried out with acoustic emission instrument and strain measurement machine with 4 channels. Since the applied stress which needs to verify an alignment should not cause any damage to the specimen, the stress level at which matrix cracking starts must be determined by the first acoustic emission. Once, we know the cracking stress from test specimen, then we can stretch all other specimen up to the stress level much below the cracking stress to check alignment. For example, if the first acoustic emission is identified at 100 MPa for the test specimen, we can apply about 50 MPa to all remaining samples to check strain variation of each surface. This low stress can be assumed not to cause any matrix cracking. If the variation of strains provided by 4 strain gauges attached to the each surface is less than 5% , then the specimen can be stretched up to the peak failure stress of composite. This will provide useful tensile stress-strain behaviour. On the other hand, if the strain varia-

tion is more than 5 % while applying stress below matrix cracking stress, then further stretching can only give more significant variation, and providing virtually useless tensile data.

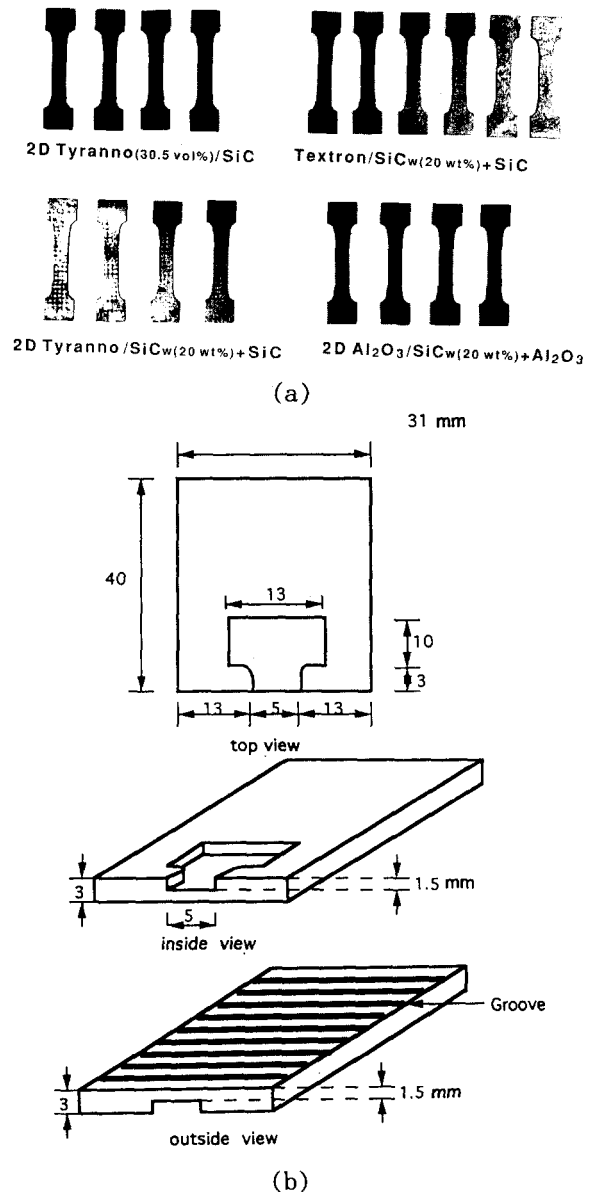


Fig. 1. Specimens machined for tension test (a) and schematic of tensile fixture (b).

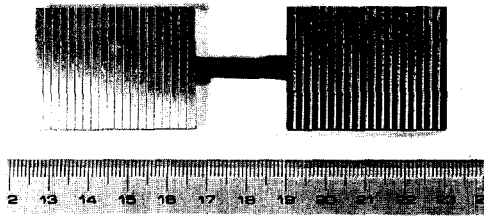


Fig. 2. Assembly of tensile specimen and fixture.

3. Results and Discussion

3.1. Mechanical properties

As-fabricated and one time impregnated composites were fractured with shear failure mode where lengthwise shear cracks propagate along the fabric/the matrix interface from the centre to the beam extremity, whereas the composites impregnated more than one time failed consistently with fibrous tensile fracture mode. Ultimate flexural strength, σ_{ub} , and proportional limit, σ_{cr} , of SiC woven laminate/SiC composites impregnated more than one time are given in Table 2. The flexural strength of matrix materials are also presented as a function of number of impregnations. Load-deflection curves of composites show a general appearance of fracture behaviour of successful composite, i.e, an initial elastic behaviour followed by an extended regime of increasing load-bearing capacity (Fig. 3). The onset of nonlinear deflection is not marked any load drop in load-deflection curve. A distinct load drop followed by other load drops after nominal peak stress could be originating

from the successive failure of the woven laminates and matrix.

Ultimate strength and proportional limit of composite increase with increasing number of impregnations. The 33 vol% fibre loaded composite shows higher relative density and strength than those of 26 vol% fibre loaded SiC composite. Comparison of strengths of matrix materials with those of fibre composites after each impregnation process reveals that the ultimate strength of composites is nearly same as the strength of matrix materials, up to 5 time impregnations. This indicates that the ultimate strength of woven laminate composite is initially determined by the matrix rather than fibre properties. The mechanical properties of both fibre composites and matrix materials given in Table 2 are again presented as a function of relative density in Fig. 4. This clearly shows that improvement of matrix density is key factor for high mechanical properties of the woven laminate composites fabricated by impregnation processes. The same tendency of increasing strength with density has also been reported in the SiC woven fabric/SiC composite fabricated by CVI process. The SiCw containing fibre composites exhibit higher value of proportional limits than those of SiC fabric/SiC composite with same relative density.

First-matrix cracking strain and elastic modulus of fibre composite given in Table 3 are obtained from flexural test performed with strain gauge attached to the centre of tensile surface under middle loading point, together with acoustic emission instrument.

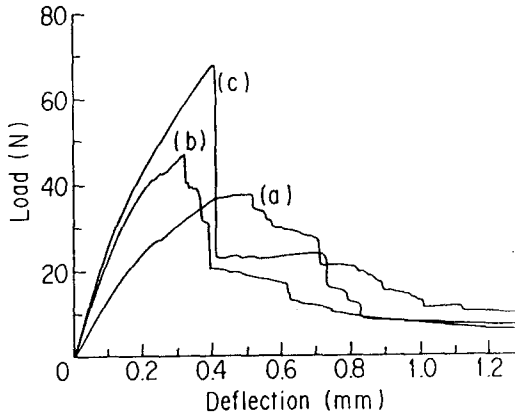


Fig. 3. Load-deflection curves of (a) two times impregnated, (b) three times impregnated and (c) five times impregnated SiC fabric (33 vol%)/SiC composites show that the proportional limit and ultimate strength of composite increase with increasing number of impregnations. Note that the load drop after peak stress is more distinct with increasing impregnation numbers.

For all of composites tested with acoustic emission technique, the first acoustic event did always occur at the proportional limit, as shown in Fig. 5. The accordance of the stress level at which first acoustic emission occurs with the proportional limit can be understood from the laminate structure of the composites. The matrix cracking in between laminae is same as crack propagation of monolithic ceramics. The cracking initiated from tensile surface is not hindered until reaching next woven laminate region, and consequently leads to the first acoustic event at the stress level accorded with the onset of nonlinear stress-strain behaviour. This results demonstrate that the flexural test equipped with strain gauge and acoustic emission instrument is very useful technique to evaluate the mechanical properties of ceramic woven fabric/ceramic matrix composites.

Table 2

Density and flexural strength of monolithic SiC and SiC woven fabric/SiC composites

Materials (Theoretical density)	SiC (3.21 g/cm ³)		SiCw+SiC (3.21 g/cm ³)		SiC fabric (26 vol %)/SiC (2.98 g/cm ³)			SiC fabric (33 vol %)/ SiCw+SiC (2.91 g/cm ³)		
	ρ (%)	σ (MPa)	ρ (%)	σ (MPa)	ρ (%)	σ_{cr} (MPa)	σ_{ul} (MPa)	ρ (%)	σ_{cr} (MPa)	σ_{ul} (MPa)
No. of impregnation										
2	64.7	116 ± 8	65.3	115 ± 3	65.7	46 ± 1	115 ± 5	67.3	51 ± 2	134 ± 5
3	65.6	146 ± 16	67.0	182 ± 14	68.2	46 ± 1	158 ± 3	70.8	60 ± 5	167 ± 2
4	66.6	163 ± 8	68.2	197 ± 10	69.2	49 ± 2	166 ± 16	72.1	88 ± 5	196 ± 5
5	—	—	68.4	211 ± 9	—	—	—	72.6	96 ± 5	216 ± 16
6	68.3	172 ± 11	—	—	71.9	61 ± 1	204 ± 10	—	—	—

The value of density was averaged from at least 10 samples. The flexural strength of composite averaged from 4 samples was obtained from load-deflection curve. A standard deviation of density is about ± 0.5. All deviations represent a standard deviation.

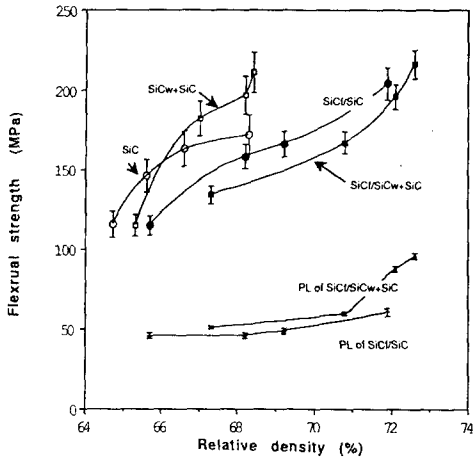


Fig. 4. Mechanical properties of SiC fabric/SiC composites and SiC matrix materials as a function of relative density show that the ultimate strength and proportional limit (PL) increase with relative density. The fibre composite containing SiCw in the matrix exhibit higher proportional limit than the composite without SiCw at same relative density.

First-matrix cracking stress is determined by the matrix failure strain and composite modulus, i.e. $E_c \epsilon_{mu}$. The SiC woven laminate (22 vol%)/SiC composite exhibits lower E_c and ϵ_{mu} values in comparison with failure strain, ϵ_m , and elastic modulus, E_m , of SiC matrix material alone (Table 3). This is likely due to the open pores in the fibre region and the yarns aligned in transverse direction to loading axis, because any fibres other than those aligned nearly parallel with the load axis can act as flaws. The reduction of E_m in the SiCw containing SiC matrix material compared with SiC alone can also be understood from the random array of SiCw. However, incorporation of SiCw into monolithic SiC increases the failure strain from 0.130 % to 0.197 %, and consequently leads to higher first-matrix cracking stress compared with the SiC matrix fibre composite, as shown in Table 2. Comparison of ϵ_{mu} and

Table 3
Mechanical properties of monolithic materials and composites

Material	T.D.	No. of	Ave.den-	$\epsilon_{mu} \times 10^3$	σ_{cr}	E	σ_{ul}
Fabric	(g/cm ³)	impreg.	sity to		(MPa)	(GPa)	(MPa)
Matrix			T.D.(%)				
—	SiC	3.21	7	68.5	1.30 ± 0.12	—	128 ± 3 166 ± 15
—	SiCw+SiC	3.21	7	68.5	1.97 ± 0.10	—	112 ± 3 220 ± 13
SiC	SiC	2.96	7	72.6	0.76 ± 0.01	77 ± 2	102 ± 4 297 ± 22
(27 vol%)							
—	Al ₂ O ₃ +SiC	3.58	7	65.7	3.20 ± 0.25	—	86 ± 2 275 ± 18
Al ₂ O ₃	SiCw+Al ₂ O ₃	3.68	4	69.0	0.90 ± 0.13	80 ± 3	95 ± 4 116 ± 16
(20 vol%)	+SiC						

The mechanical properties were obtained from stress - strain curve. First - matrix cracking stress was identified by the first acoustic event during loading. The data from four specimens were averaged for each value of mechanical property.

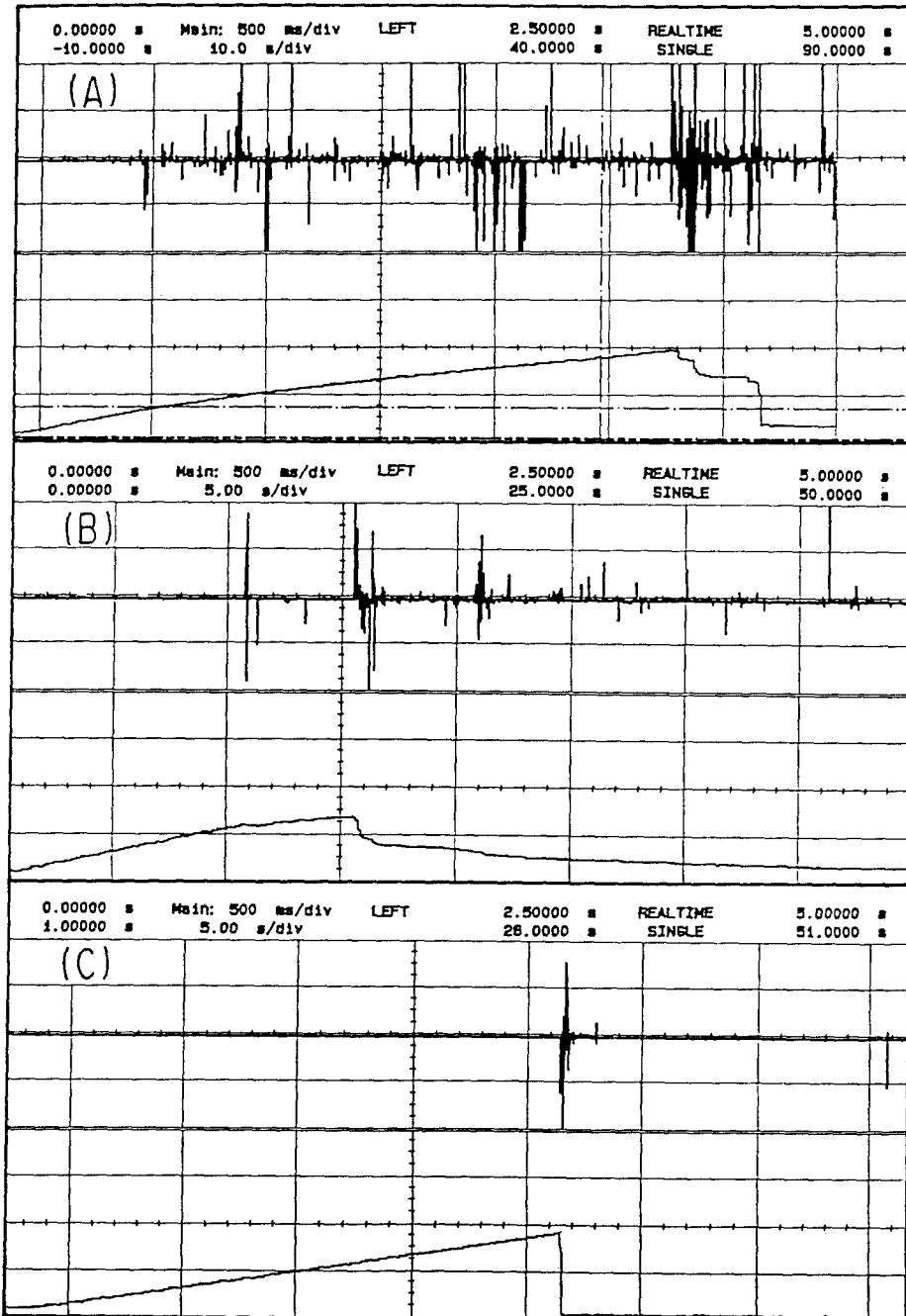


Fig. 5. Load and acoustic events as a function of testing time during flexure test of (a) SiC fabric (27 vol %)/SiC composite, (b) Al_2O_3 fabric (20 vol %)/SiCw + Al_2O_3 + SiC composite and (c) Al_2O_3 + SiC matrix materials show that the first acoustic event occurs at the proportional limit.

E_c of the four times impregnated Al_2O_3 woven laminate composite with those of SiC fabric composite shows that the higher ε_{mu} of Al_2O_3 fabric composite can compensate the lower E_c , and results in nearly same value of first-matrix cracking stress.

The flexural strength of 297 ± 22 MPa obtained from SiC fabric (22 vol%)/SiC composite is comparable with the values (288-310 MPa) of same type composite with 81-90 % of relative density fabricated by CVI process, despite of lower density (72.6 % of T.D) and fibre loading.

3.2. Ratio of ultimate flexural strength to ultimate tensile strength

The ultimate flexural and tensile strength measurements on the ceramic matrix fibre composites give quite different results, as evident in Tables 4 and 5. The ratio between the flexural and tensile strengths is generally around 1.5 for the uniaxially aligned fibre composites. However, the Weibull modulus of the composite measured using these two types of test was found to be identical [7], implying that the distribution in strength values is not a function of testing methods. The higher strength obtained by flexural test compared with the tensile strength can be explained by several reasons. In addition to the neutral axis shift to compression side due to both the initiation of matrix cracking and the growth of pre-existing cracks at tensile face, the application of Weibull statistics can allow us to find the ratio of the strengths of a bar of material breadth b ,

depth d and length l in tension and in bending.

Conventionally, for monolithic ceramics, the ratio of the bending strength (σ_b) to the tensile strength (σ_t) can be estimated by the relative stressed volumes in tension (V_t) and in flexure (V_b) and Weibull modulus of materials (m) [11,12] In order to calculate the probability of survival in tension or in bending, we first need to know the variation of stress with volume. The survival probability can be again expressed by.

$$P_s = \exp \left[- \int_v \left(\frac{\sigma}{\sigma_0} \right)^m dV \right] \quad (1)$$

In tension we assume that whole bar is uniformly stressed, and thus the integral is simply $V(\sigma_t/\sigma_0)^m = bdl(\sigma_t/\sigma_0)^m$, so the probability of survival in tension at stress σ_t is, from Eq (1)

$$P_s(\sigma_t) = \exp \left[- bld \left(\frac{\sigma_t}{\sigma_0} \right)^m \right] \quad (2)$$

In 3-point bending, stress varies with distance from the point of maximum stress σ_{3b} on the surface opposite the central knife-edge, both with distance into the beam and along the beam. Thus $\sigma = \sigma_{3b}(1-2d'/d)(1-2l'/l)$ where d' is the distance from the surface into the beam and l' is the distance from the centre along the beam [11]. At the neutral axis of the beam, half of the specimen under a compressive stress can be neglected for the purpose of tensile failure. The integral in the equation (1) for 3-point bending is,

Table 4
Tensile properties of the composites

Material		T.D.	No. of	Ave.den-	$\varepsilon_{mu} \times 10^{-3}$	σ_{cr}	E	σ_{ul}
		(g/cm ³)	impreg.	sity to		(MPa)	(GPa)	(MPa)
Fabric	Matrix			T.D(%)				
Ty SiC (30 v/o)	SiC	2.93	5	76.6	0.37 ± 0.02	38 ± 2	102 ± 4	97 ± 4
Ty SiC (30 v/o)	SiCw + SiC	2.93	5	73.5	0.51 ± 0.01	51 ± 2	100 ± 2	90 ± 5
Almax (16v/o)	SiCw + Al ₂ O ₃ + SiC	3.59	4	73.8	0.27 ± 0.02	25 ± 5	92 ± 4	43 ± 4
Tex (14 v/o)	SiCw + SiC	3.17	5	70.0	0.73 ± 0.16	118 ± 8	160 ± 4	374 ± 5

The mechanical properties were obtained from stress-strain curve. First-matrix cracking stress was identified by the first acoustic event during loading. The data from four specimens were averaged for each value of tensile property.

Table 5
Comparison of the ultimate flexural and tensile strengths of composites

Woven fabric composites		Flexural σ_{ul}	Tensile σ_{ul}
		(MPa)	(MPa)
Fabric	Matrix		
Tyrano SiC	SiC	297 ± 22	97 ± 4
Almax Al ₂ O ₃	SiCw + Al ₂ O ₃ + SiC	116 ± 16	43 ± 2

$$\left(\frac{\sigma_{3b}}{\sigma_0}\right)^m b \int_0^{d/2} (1 - 2d'/d)^m dd' 2 \int_0^{l/2}$$

$$(1 - 2l'/l)^m dl' = \frac{bd}{2(m+1)^2} \left(\frac{\sigma_{3b}}{\sigma_0}\right)$$

Then, the probability of survival in three point bending is

$$P_s(\sigma_{3b}) = \exp\left[-\frac{bd}{2(m+1)^2} \left(\frac{\sigma_{3b}}{\sigma_0}\right)\right] \quad (3)$$

For 4-point bending, the stress within the inner span does not vary along the beam,

and changes only with distance into the beam, and most of failure occurs within the inner span region. Thus, the integral in equation (1) for 4-point bending can be assumed as [12]

$$\begin{aligned} &\left(\frac{\sigma_{4b}}{\sigma_0}\right)^m bl \int_0^{d/2} (1 - 2d'/d)^m dd' \\ &= \frac{bd}{2(m+1)} \left(\frac{\sigma_{4b}}{\sigma_0}\right)^m \end{aligned}$$

The probability of survival in 4-point bending is

$$P_s(\sigma_{4b}) = \exp\left[-\frac{bd}{2(m+1)}\left(\frac{\sigma_{4b}}{\sigma_0}\right)\right] \quad (4)$$

We can now compare the maximum failure stress in each case by equating the cumulative probabilities of failure, which is same as equating the probabilities of survival shown in Eqs (2), (3) and (4). The ratio of the failure stress in tensile test to the maximum failure stress in 3-point bending or in 4-point bending can be given as

$$\sigma_t = \frac{\sigma_{3b}}{[2(m+1)^2]^{1/m}} = \frac{\sigma_{4b}}{[2(m+1)]^{1/m}} \quad (5)$$

This shows why $\sigma_t < \sigma_{4b} < \sigma_{3b}$.

Assuming that (i) a fibre composite is only an array of fibres dead loaded at the both ends in which the heavily microcracked matrix is assumed not to contribute to the ultimate strength of composites at ultimate failure, and (ii) the load from a broken fibre is assumed to transfer to the entire length of the all remaining fibres, i.e, equal load sharing. Then, the strength of fibre bundles are sensitive to the number of fibres in the bundle only for very small numbers of fibre. The accuracy of bundle strength will be within $\sim 10\%$ for 10 fibres and $\sim 1\%$ for 100 fibres, thus in practice the number of fibres in the bundle will not be important. However, the length of fibres and the strength distribution along fibre will be important. The ratio of σ_b to σ_t for equal volume specimens is then proportional to $2(m+1)^{1/m}$, as predicted from bending case. The ratio σ_b/σ_t can be obtained by equating the cumulative probability of failure of each case, as expressed in Eq (1) [13]. Thus,

$$\frac{\sigma_b}{\sigma_t} = \left[2(m+1)\frac{l_t}{l_b}\right]^{\frac{1}{m}} \quad (6)$$

where l_b and l_t are the length of fibre under stress in flexural and tension tests, respectively. In the case of Nicalon fibre ($m = 4$) composite with assumption of same stressed length ($l_b = l_t$), the ratio σ_b/σ_t will be ~ 1.58 for the uniaxially aligned fibre composites. For the 2D fibre composites, the ratio will be increased in comparison with the value of uniaxial fibre composites, since the fibres aligned at 90° with respect to the axis of tensile stress cannot contribute to the improvement of the ultimate strength. Thus the data shown in Table 4 could be understood in this respect.

4. Conclusions

The mechanical properties of newly developed SiC fabric/SiC and Al_2O_3 fabric/ Al_2O_3 + SiC composites were successfully characterized by the both flexure and tension tests. The 7 times impregnated SiC woven fabric (22 vol%)/SiC composite exhibits 77 MPa of first-matrix cracking stress and about 300 MPa of ultimate flexural strength which are comparable to the values obtained from the SiC woven laminate composite fabricated by CVI process. The higher first-matrix cracking stress of SiCw dispersed SiC matrix fabric composite was attributed to the increment of matrix failure strain due to the addition of SiCw. A higher ultimate flexural strength in comparison with ultimate tensile strength for the same fibre composite

could be explained by considering the relative stressed volumes in tension and in flexure using Weibull statistics.

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