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Journal of Nuclear Materials 278 (2000) 127–135

journal of nuclear materials

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Joining of SiC/SiC$_f$ ceramic matrix composites for fusion reactor blanket applications

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Received 24 June 1999; accepted 15 November 1999

Abstract

Using a preceramic polymer, joints between SiC/SiC$_f$ ceramic matrix composites were obtained. The polymer, upon pyrolysis at high temperature, transforms into a ceramic material and develops an adhesive bonding with the composite. The surface morphology of 2D and 3D SiC/SiC$_f$ composites did not allow satisfactory results to be obtained by a simple application of the method initially developed for monolithic SiC bodies, which employed the use of a pure silicone resin. Thus, active or inert fillers were mixed with the preceramic polymer, in order to reduce its volumetric shrinkage which occurs during pyrolysis. In particular, the joints realized using the silicone resin with Al–Si powder as reactive additive displayed remarkable shear strength (31.6 MPa maximum). Large standard deviation for the shear strength has nevertheless been measured. The proposed joining method is promising for the realization of fusion reactor blanket structures, even if presently the measured strength values are not fully satisfactory. © 2000 Elsevier Science B.V. All rights reserved.

PACS: 81.05.Mh; 81.20.Vj

1. Introduction

SiC/SiC$_f$ ceramic matrix composites (CMCs) are an attracting material for fusion reactors because of their good mechanical properties at high temperature, low chemical sputtering, high oxygen gettering and very low activation at short and medium terms [1,2]. Nevertheless, there are some critical issues connected to the long production time, their high porosity and the capability to accomplish complex geometries that still need to be solved. Moreover, since these materials can only be produced in limited sizes and shapes, in order to assemble a complete blanket system, a suitable method of joining SiC/SiC$_f$ components is required.

To develop a joining technology compatible with the stringent low activation criteria, the needed low fabrication temperature and with the in service loads estimated within the DEMO blanket design activities [3], a task was launched in the framework of the EURATOM Long-Term Fusion Technology Program. The basic idea of the joining method investigated relies on the extension to SiC/SiC$_f$ 2D and 3D composites of the promising results obtained joining sintered α-SiC and reaction bonded SiC monolithic bodies with a preceramic polymer [4–6]. Joining is realized through the application of a layer of a Si-based preceramic polymer which, during pyrolysis at high temperature, transforms into a ceramic material, yielding an adhesive bonding with the SiC bodies. It is well known that preceramic polymers can be used successfully to produce a large range of ceramic products, including fibers, films, foams and fiber-reinforced composites [7,8], but only recently their potential for producing ceramic joints has been investigated [9–12]. One of the major problems encountered is the

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PII: S0022-3115(99)00268-8

volumetric shrinkage of the bonding layer occurring during the polymer-to-ceramic conversion, which reduces the strength of the joints [12]. Thus, several modifications of the basic 'gluing' method have been proposed, including the use of inert or reactive powder fillers [5,9–11] or fibers [10], the application of pressure during pyrolysis [4,12], and the performing of multiple impregnation/pyrolysis cycles [10].

Despite the not yet optimal strength values achieved so far, joining of SiC/SiC$_f$ composites for nuclear applications with preceramic polymers seems to be one of the most promising methods [13] among the various ones proposed in the literature (like brazing [14], reaction bonding [15,16], the use of glassy interlayers [17,18], the use of in situ displacement reactions [19,20]).

2. Experimental

2.1. Materials

The experimental work was carried out using two SiC/SiC$_f$ composites produced by the European company SEP Division of Snecma (Saint-Medard-en-Jalles, France). They consist of a bi-directional cloth (CERASEP N2-1) and of a tri-dimensional weave (CERASEP N3-1) of Nicalon fibers densified by chemical vapor infiltration (CVI) and finally coated by chemical vapor deposition (CVD) with methyltrichlorosilane and hydrogen at about 1000°C (SEP proprietary process). The general properties of the composites can be found in Ref. [21].

The preceramic polymer used for the joining experiments was a methyl-hydroxyl-siloxane (SR350, General Electric Silicone Products, Waterford, NY). Its pyrolysis in inert atmosphere yields a $Si_xO_yC_z$ ceramic, with a ceramic yield of about 85% [22,23]. The SR350 resin was dissolved in absolute ethanol to yield solutions of different concentrations. In order to control the rheology of the polymer solution, in some instances a catalyst was introduced (amino-propyl-triethoxy-silane). In fact, a high viscosity would yield thick joints with the consequent development of voids and cracks in the ceramic layer (due to the above-mentioned polymer-to-ceramic conversion which occurs with a large shrinkage during pyrolysis), while a low viscosity would lead to very thin and inhomogeneous joints. Moreover, different routes were tested in order to try to improve the strength of the joints: the introduction in the bonding layer of inert powder (SiC nano-powder, mean diameter = 20–40 nm, produced by ENEA by laser pyrolysis), of active powder (Si powder, mean diameter < 10 μm; Si nano-powder, mean diameter = 20–40 nm; 88Al–12Si powder, mean diameter < 15 μm), or of ceramic fibers and cloths (plane weave carbon or Nicalon (SiC) fiber cloth, carbon or Nicalon fibers impregnated with SR350 resin, carbon felt impregnated with SR350 resin). Powders were added in various amounts to reduce the shrinkage, while ceramic fibers and cloths were added to try to realize non-brittle (composite) joints.

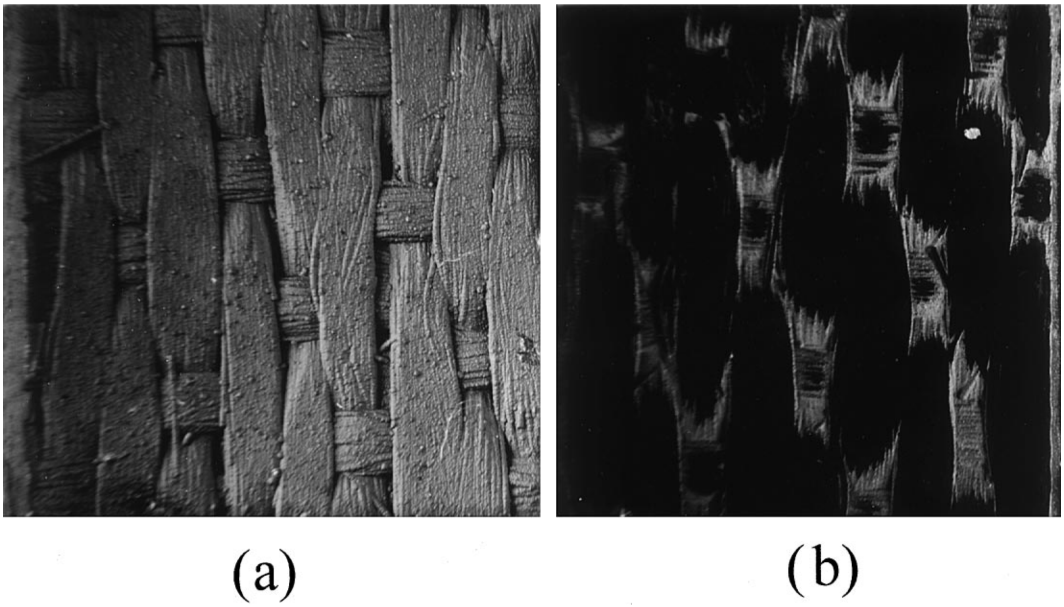

Fig. 1. (a) Stereo-micrograph of an as-received 3D composite specimen, (40×). (b) Stereo-micrograph of a polished 3D composite specimen, (40×). Flat areas appear black.

Reinfiltration experiments were also performed, using a diluted polymer solution without additives. After the infiltration in a vacuum container, the specimens were pyrolyzed, and the cycle was then repeated a few times.

2.2. Processing of the joints

Composite samples were cut into $10 \times 8 \times 3$ mm^3 bars from longer specimens. The surface morphology of the 2D and 3D composites was rather complex (roughness Ra ~15–25 μm, average peak-to-valley height Rz ~100–150 μm). Direct joining of as-received specimens was never successful, because the rough surfaces contain cavities in which the preceramic polymer is subjected to a large shrinkage upon pyrolysis, and the resulting defects in the ceramic layer behave as crack initiation sources [5].

In order to improve the surface quality, the specimens were mechanically polished on one side, using SiC and diamond paper, and ultrasonically cleaned in acetone prior to joining. After this procedure, the total roughness of both materials (measured on an area a few square millimeters wide) was greatly reduced. Generally speaking, it appears that the final roughness achievable for a 2D composite was lower than the one for a 3D composite. After grinding, the SiC/SiC$_f$ composite's surface was comprised of flat areas (a few hundred μm^2 wide) and of deep 'valleys' where the fibers are interwoven. The surface roughness, measured only on the flat areas where joining mainly takes place, was of the order of 0.25–0.5 μm. It has to be pointed out that the SiC/SiC$_f$ composites have a thick SiC over-coat (>100 μm) deposited by CVD, which was not completely removed by the mechanical surface preparation. Thus the fibers were not exposed and did not participate to the joint formation.

In Fig. 1(a) and (b) are shown optical stereo-micrographs of as-received and polished 3D composites. In Fig. 2 are reported the curves obtained using a SURTRONIC 3+ profilometer. It can be observed that the real area of contact (flat zones) is smaller than the nominal one, typically of the order of 60–80% for the 3D composite specimens.

The viscous solutions were homogeneously applied, using a spatula, to the surface of the specimens to be joined, and the samples were overlapped to establish the joint. Subsequently, the samples were loaded with a moderate axial pressure (0.01–0.02 MPa) and heated at 100–150°C for 2 h in air, in order to achieve a uniform thickness of the joining layer favored by the melting of the silicone resin, and to allow the complete crosslinking of the preceramic polymer. The control of the rheological characteristics of the joining material (silicone resin plus fillers) at this stage seems to be crucial for obtaining a homogeneous distribution of the bonding layer on the composite's surface, and thus better mechanical properties for the joint. Depending on the joining material used (pure preceramic polymer, polymer plus powder fillers, polymer plus fibrous fillers), the joint thickness and viscosity of the solution was varied by changing the solution concentration, the curing temperature, the amount of catalyst added and the axial load applied. When using C or SiC fiber bundles, they were aligned in the direction parallel to the shear load direction; cloths, felts and fibers were pre-impregnated with a diluted solution of SR350.

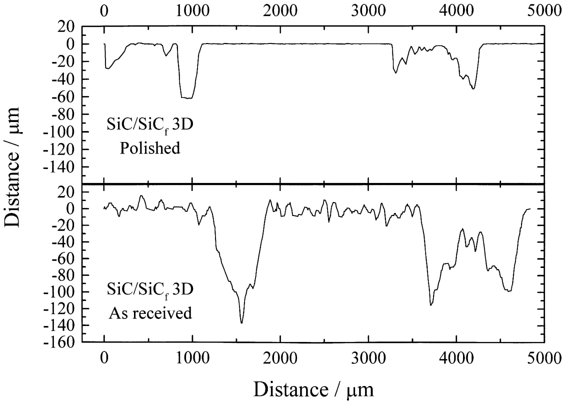

Fig. 2. Surface linear profile of as-received and polished 3D composite specimens.

The samples were then heated at temperatures between 1000°C and 1400°C, with dwelling times ranging from 1 to 8 h, in different atmospheres (flowing air, nitrogen (99.99%) or argon (99.99%)). The heating and cooling rates were rather slow (2°C/min) in order to minimize residual stresses due to the (limited) thermal expansion mismatch.

2.3. Characterization of the joints

The joint quality was determined by microstructural examination using a scanning electron microscope (SEM) and by shear tests, performed following a modification of the ASTM D905-89 test procedure. The crosshead speed was 1 mm/min. Even if the presence of a pure shear stress field was not assured by this method of testing, it was however a suitable means for a comparative evaluation of the different specimens. The basic principle of the test is shown in Fig. 3(a) and (b). The value of the shear strength was the main parameter considered for ranking the different joining materials and parameters used. Each data point represents the average value of at least 5 (in most cases at least 10) individual tests. The strength values were computed using the nominal surface area of the specimens, because of the difficulty to estimate the exact contact zone for each sample. In some cases the samples were used again

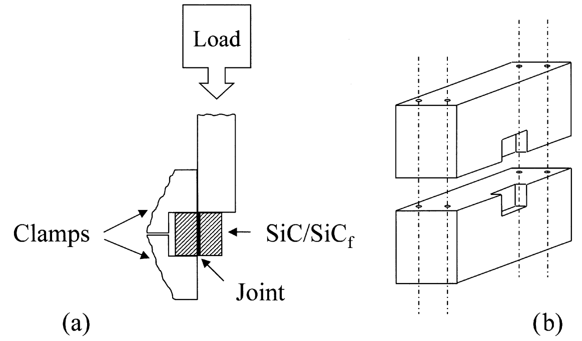

Fig. 3. (a) Basic principle of the shear test procedure adopted (following ASTM D905-89). (b) Clamping devices used.

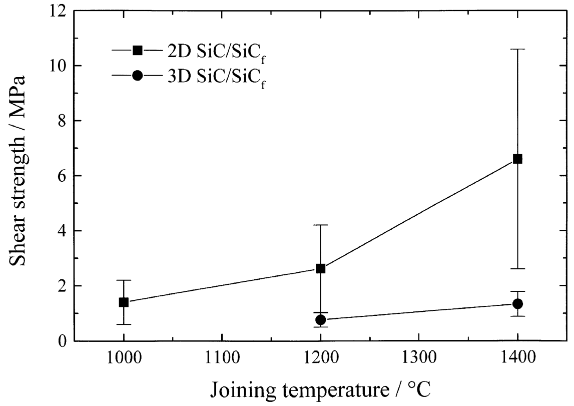

Fig. 4. Shear strength as a function of pyrolysis temperature, for joints realized with 2D and 3D composites using pure SR350 preceramic polymer.

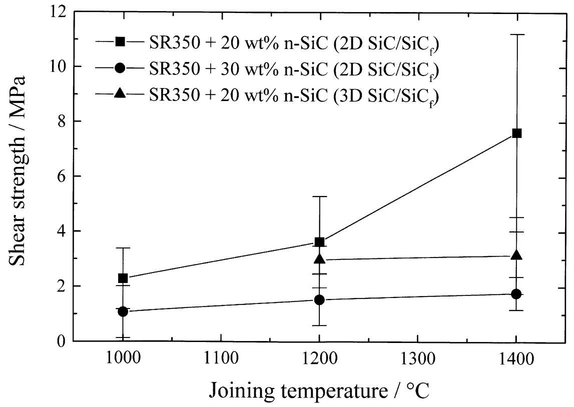

Fig. 5. Shear strength as a function of pyrolysis temperature, for joints realized with 2D and 3D composites using SR350 and *n*-SiC powder.

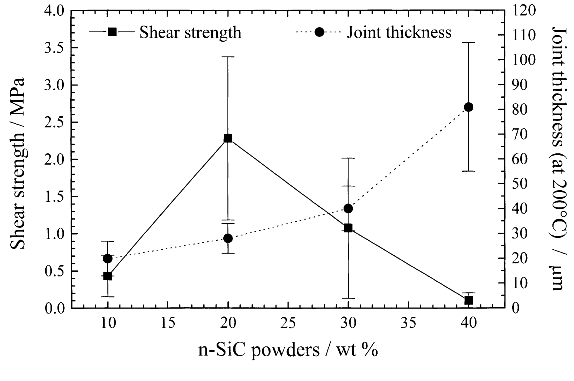

Fig. 6. Shear strength and joint thickness (before pyrolysis) as a function of inert filler (*n*-SiC) content, for joints realized with 2D composites (samples pyrolyzed at 1000°C).

after mechanical testing, removing the residues of the joining layer with a careful polishing.

The thickness of the joint after the crosslinking step was measured by using a micrometer, since its value and uniformity can give information on the chances of success of the joint [4,12]. Generally speaking, at this stage joints obtained from solutions of pure SR350 or SR350 plus powders had a thickness of 20–100 μm; after pyrolysis, the thickness drastically decreased, because of the polymer-to-ceramic transformation [4–8]. After pyrolysis, the joint thickness can be measured accurately only by microscopy, usually being of the order of 2–10 μm.

The microstructure of the joining material was investigated by X-ray diffraction (using a Philips 1730/1820 diffractometer equipped with a thin film attachment (Cu-Kα radiation, glancing angle = 0.05°)).

3. Results

3.1. Mechanical properties

3.1.1. Joining using a pure preceramic polymer

In Fig. 4 is shown the shear strength of joints realized with 2D and 3D composites using a pure preceramic polymer (SR350 silicone resin). The joint thickness (measured after the crosslinking step at 200°C, but before pyrolysis) was in the range 20–30 μm. The dwelling time at pyrolysis temperature (argon atmosphere) was 2 h. Joints of 2D composites possess a higher strength than 3D ones, which increases with increasing pyrolysis temperature.

3.1.2. Joining using a preceramic polymer and an inert filler

In Fig. 5 is shown the influence of the pyrolysis temperature on the joints (2D and 3D composites, dwelling time 2 h); the shear strength increases with increasing processing temperature, and joint strength is higher for 2D specimens than for 3D ones. In Fig. 6 is reported the shear strength and thickness (measured at 200°C, before the pyrolysis step) of joints performed with 2D SiC/SiC$_f$ composites using SR350 preceramic polymer and inert fillers (nano-SiC powder), as a function of inert filler content. The joints were performed by pyrolysis for 2 h at 1000°C in argon.

3.1.3. Joining using a preceramic polymer and an active filler

In Fig. 7 is reported the shear strength of joints performed with 2D and 3D composites using SR350 preceramic polymer and active fillers (88Al–12Si powder), as a function of processing temperature (dwelling time 1 h, argon atmosphere). Two different compositions were tested (weight ratio SR350/Al–Si = 1.67 and SR350/Al–Si = 0.78); the joint thickness (measured after the crosslinking step at 200°C, but before pyrolysis) was about 30 μm. While it appears that no major difference is found for 2D composites when joining is performed using inert or active fillers, joints with 3D composites display a substantial strength increase, which depends both on the filler content and the processing temperature. An increase of the dwelling time, from 1 to 8 h did not yield any increase of the measured strength, as well as performing the joining process in nitrogen or in air atmosphere. Some samples were joined using a mixture (of variable composition: SR350 40 wt%, Al–Si = (60 − x) wt%, n-SiC = x wt%) of active (Al–Si) and inert (n-SiC) fillers. The data obtained for pyrolysis at 1200°C in Ar, dwelling time 1 h, show that a substantial decrease in joint strength occurred for n-SiC contents above 6 wt%, and for lower amounts of nano-silicon carbide no improvements over the use of only active fillers was observed. The use of Si powder as active filler did not allow to obtain joints possessing high shear strength (maximum strength = 3 MPa), and XRD investigations showed that Si remained largely unreacted even after processing at 1400°C.

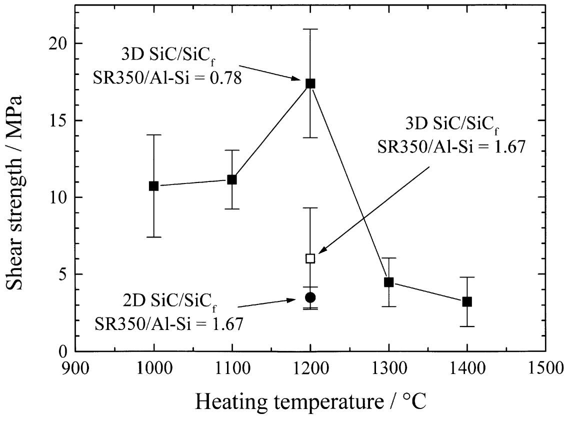

Fig. 7. Shear strength as a function of pyrolysis temperature, for joints realized with 2D and 3D composites using SR350 and Al–Si powder.

3.1.4. Joining using a preceramic polymer and ceramic fibers

In Table 1 is reported the shear strength of joints performed with 2D SiC/SiC$_f$ composites using SR350 preceramic polymer and ceramic fibers (Nicalon and C fiber bundles, Nicalon and C 2D cloths, and C felt). The joints were performed by pyrolysis at 1200°C for 1 h in Ar. The joint thickness (measured after the crosslinking step at 200°C, but before pyrolysis) was in the range 150–300 μm. The stress–strain curves relative to the shear strength testing demonstrated the occurrence of non brittle fracture, but the joint strength was not adequate. SEM investigations showed that, even when the fibers, cloth, or felt were infiltrated with SR350 and pyrolyzed prior to be inserted in the joining layer, the resulting joints contained many voids. No benefits were found increasing the pyrolysis temperature or changing the orientation and content of fibers.

3.1.5. Reinfiltration of joints

In Fig. 8 is reported the shear strength of joints performed with 3D composites using SR350 preceramic polymer and 88Al–12Si powder (SR350/Al–Si = 0.78). After pyrolysis at 1200°C for 1 h in Ar, the specimens were infiltrated (with the application of vacuum) using a diluted solution of pure SR350 and then pyrolyzed again at 1200°C. This cycle was repeated up to 4 times. The joint thickness (measured after the crosslinking step at 200°C, but before pyrolysis) was about 30–50 μm. A strong increase of the shear strength with the number of reinfiltration cycles, as well as a decrease of the data scattering, can be observed. The lower strength value for specimens not reinfiltrated, with respect to data displayed in Fig. 7, is probably due to a different thickness (larger) of the initial joining layer.

Table 1

Mean and maximum shear strength values (MPa) of 2D SiC/SiC$_f$ specimens joined using SR350 resin and ceramic fibers

SR350 + Nicalon fibers	SR350 + C fibers	SR350 + Nicalon cloth	SR350 + C cloth	SR350 + C felt
1.0 ± 0.7 Max value = 2.7	0.6 ± 0.4 Max value = 1.4	0.8 ± 0.2 Max value = 1.0	0.3 ± 0.1 Max value = 0.4	1.0 ± 0.8 Max value = 3.4

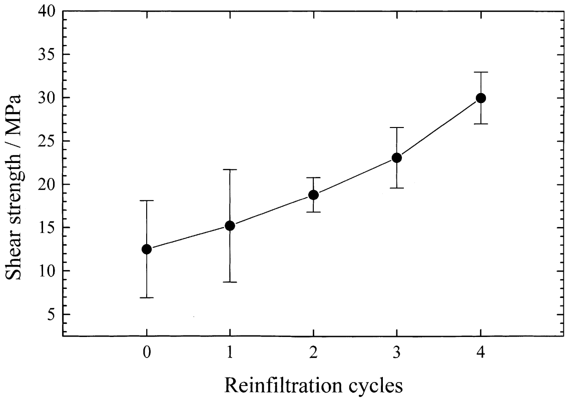

Fig. 8. Shear strength as a function of the number of reinfiltration cycles, for joints realized with 3D composites using SR350 and Al–Si powder.

3.2. Microstructural and morphological analysis

3.2.1. Microstructural analysis

Glancing angle X-ray diffraction analyses were performed, after mechanical testing, on the surface of specimens joined using SR350 and Al–Si powder, to investigate which phases formed upon pyrolysis. In particular, it was interesting to monitor the eventual disappearance of metallic Al from the joining material. For the sake of brevity, the results will only be summarized here.

The analysis showed that, when the pyrolysis was performed in Ar, Al reacted with the silicone preceramic polymer yielding aluminum oxide, through the reaction

$$4\,\mathrm{Al} + 3\,\mathrm{SiO_2} \rightarrow 2\,\mathrm{Al_2O_3} + 3\,\mathrm{Si}, \qquad (1)$$

while the Si thus produced reacted with the carbon present in the preceramic polymer, to form silicon carbide, according to the reaction

$$\mathrm{Si} + \mathrm{C} \rightarrow \mathrm{SiC}. \qquad (2)$$

Differential thermal analysis (DTA) demonstrated that the above reactions happen only after the melting of the Al–Si powder (580°C), and this is beneficial because it ensures a more homogeneous distribution of the reaction products within the joining layer, instead of concentrating them in scattered particles. Reaction 2 seems to occur significantly only for pyrolysis temperatures above 1100°C. Only for pyrolysis temperatures of 1400°C both Al and Si crystallites disappear from the XRD spectra; however, the amount of metallic Al and Si remaining in the joint material, at 1200°C, is very limited. An increase of the pyrolysis time at 1200°C (from 1 to 8 h) was not successful in reducing the amount of unreacted phases. In general, the pyrolysis cycle at 1400°C did not seem to cause problems related to the degradation of the SiC fibers in the composite. No formation of brittle $\mathrm{Al_4C_3}$ was observed, because of the well-known inhibiting effect of Si in this respect. The use of a silicone resin is advantageous, because the removal of metallic Al from the joining layer (through reaction with the oxygen-containing preceramic polymer) is expected to avoid the decrease of the mechanical properties of the joint with increasing testing temperature. This has been in fact observed for joints obtained using Al–Si powder and a different preceramic polymer, which did not contain oxygen [24]. In conclusion, the joining material in the case of using the Al–Si active filler and processing in Ar (at 1200°C) is comprised of $\mathrm{Al_2O_3}$ and SiC crystallites (with an average dimension of 25–30 nm) embedded in an amorphous phase (silicon oxycarbide), with traces of Al and Si.

Processing in air allowed to completely eliminate the presence of Al after 8 h of pyrolysis at 1200°C; however Si remained largely unreacted, and such conditions are probably too severe for the composite material itself. Processing in nitrogen formed nitrided species (AlN), which are not compatible with the low activation requirement for nuclear fusion materials. It has to be noted that, even if aluminum is not a low activation material, it can be tolerated in the small quantities used for the joining of the blanket CMC structural components. In fact it has been proposed as a candidate for deuterium permeation barriers deposited on $\mathrm{SiC/SiC_f}$ composites [25].

The joining material, in the case of joints preformed using pure SR350 preceramic polymer, was an amorphous silicon oxycarbide ceramic [5,6], while SiC crystallites were obviously present in the joints obtained from mixtures of SR350 and SiC powder.

3.2.2. SEM investigations

In Fig. 9(a)–(c) are shown SEM micrographs of the cross-section of a 3D $\mathrm{SiC/SiC_f}$ composite joined using SR350 and Al–Si powder (1 h 1200°C, Ar, SR350/Al–Si = 0.78), and reinfiltrated 4 times. The average joint thickness appears to be about 40 μm, but a large variation in the thickness can be observed in the regions where the surface of the substrates was not flat. The pyrolysis products appear to be continuous, rather dense and well adherent to the $\mathrm{SiC/SiC_f}$ parts, and the craters in the composite's surface seem to have been almost completely filled by the joint material. However, where the joint thickness is large (> 200 μm), cracks appear in the joining layer. Similar investigations performed on specimens not reinfiltrated showed that the joints appeared to contain a larger amount of porosity. The amount of porosity in the joints was not easily quantifiable, because it consisted mainly of a network of very small pores, as often happens in polymer-derived ceramics [8], more than sizable scattered voids in the layers. Preliminary X-ray microanalysis investigations showed a homogeneous distribution of Al throughout

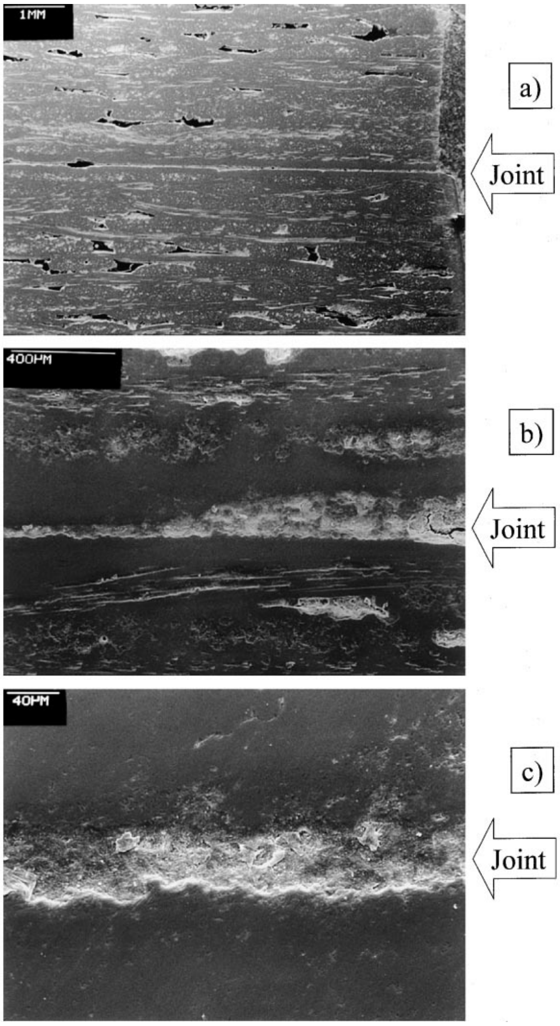

Fig. 9. (a) SEM micrograph of the cross-section of a joint realized with a 3D composite, using SR350 and Al–Si powder and reinfiltrating 4 times. (b) Detail of the joint region, showing the presence of cracks in the joining material in an area of large joint thickness. (c) Detail of the joint region, showing how the surface roughness is well filled by the joining material.

the joint thickness, as expected because of the above-mentioned reaction mechanism (i.e. at first melting of the Al–Si filler, followed by the reaction with the silicone resin). No macroscopic reaction layers were visible at the interface SiC/SiC$_f$ – joint, and thus we have reason to believe that, also in the case of Al–Si active filler, adhesion is the basic bonding mechanism, as already observed when joining with a pure preceramic polymer [5,6]. For not reinfiltrated samples, the joining material did not enter into the composite and did not fill the inner (sub-surface) open porosity.

Observation of the fracture surfaces revealed that failure was nearly always cohesive (it occurred mostly in the joint layer, and not at the interface between the joint material and the SiC/SiC$_f$ composites), and the pyrolysis residues covered homogeneously most of the surface of the specimens.

4. Discussion

Considering that the polymer-to-ceramic conversion during pyrolysis occurs with a large volume shrinkage [7,22,23], it is evident from the surface analysis (Figs. 1 and 2) that as-received composites do not possess a surface texture suitable for joining with a pure preceramic polymer. The polymer would fill the deep craters present on the surface, but the pyrolysis residues would be largely cracked, thus strongly reducing the joint strength. The difference in measured strength between polished 2D and 3D composites (see Fig. 4), which do not differ in chemical composition, suggests that even after flattening the surface of the specimens the morphology plays an important role in the joining process. This is one of the main reasons why joints between SiC/SiC$_f$ composites made with a pure preceramic polymer displayed much lower strength with respect to joints between monolithic SiC specimens [4–6].

Using inert fillers as a way of reducing the shrinkage afforded very limited improvement to the shear strength for 2D composites joined at low temperatures (see Fig. 5), while the strength for 3D composites more than doubled, but remained unacceptably low. This finding is an indication that the effect of the filler varies depending on the surface morphological characteristics of the specimens. After an initial increase, the shear strength decreased with increasing joint thickness and with reducing the polymer content (see Fig. 6), as already observed [4] and as can be expected because of increasing residual stresses [26] and residual porosity in the joints. Joints obtained using a pure preceramic polymer have been demonstrated to contain little or no residual stresses [6], but a composite material (SiOC plus fillers) can have a rather different CTE than SiC/SiC$_f$ [27]. However, no evidence of delamination or cracks running along the SiC-joint interface was found by SEM investigations.

Generally, it appears that increasing the processing temperature improved the strength of the joints, certainly because of the decreasing of the final joint thickness due to the proceeding of the pyrolysis reaction, with a concurrent increase in the density of the ceramic product. When using the Al–Si active filler, however, a strong decrease in the joint strength was observed for specimens processed above 1200°C (see Fig. 7). This is attributed to the formation of a large amount of open porosity at higher pyrolysis temperatures, as observed when studying bulk specimens pyrolyzed at 1400°C [27]. The amount of active filler present appears to play an important role in the development of strong joints and, again, a dependence on the type of ceramic composite used (2D or 3D) was observed. As mentioned before, the choice of an active filler which melts before reaction (Al–Si powder) is beneficial because of the good distribution of the products in the films. Moreover, the reaction products are crystalline (Al_2O_3, SiC), and reduce the amount of silicon oxycarbide glassy phase, whose stability under irradiation is questioned [28].

The introduction of inert and active fillers allowed to greatly decrease the shrinkage occurring during the pyrolysis of the preceramic polymer, thus effectively reducing defects in the layer. In fact, the joint thickness observable in the SEM micrographs – Fig. 9(b) and (c) – is very similar to the initial thickness of the joint before pyrolysis, while the thickness of joints made of pure preceramic polymer is about 2–5 μm after pyrolysis [5]. However, a few reinfiltration cycles were necessary for obtaining a significant improvement in the shear strength values, because they eliminated the residual porosity present in the joints. The difference in effectiveness between inert and active fillers might be related to the different distribution of the filler within the joining layer (scattered particles in the first case, homogeneous distribution of reaction products in the second one). The rather large scattering present in the mechanical tests data has different causes, like the possible variation of joint thickness among different samples (due to the manual alignment and application of the joining paste), the different area which is actually available for the joint (mainly the flat regions), the amount of defects in the joint layer (which depends both on the degree of polishing and on the original surface morphology of the specimens). From the data presented, which cover various materials and experimental conditions, it is evident that the joint strength depends simultaneously on a large number of variables (composition of the joining material, pyrolysis atmosphere, pyrolysis temperature and dwelling time, morphology of the SiC/SiC$_f$ composite, preparation of the surface, etc.) whose optimization will require more experimental work.

5. Conclusions

The proposed joining technique, which employs a preceramic polymer and fillers, seems in principle suitable for joining large surfaces of SiC/SiC$_f$ composite, like those that will need to be fabricated for the breeding blanket of fusion reactors. In this way it is, in fact, possible to obtain joints free of residual stresses, with thermo-mechanical properties similar to those of the SiC/SiC$_f$ composites, and in principle possessing the low activation requisite.

Nevertheless, adhesive joining with preceramic polymers has some disadvantages, like the need of grinding the surfaces of the composites, and the residual porosity of joints at the interfaces. However, by reinfiltrating a few times the joined samples using a pure preceramic polymer solution, the joint strength can be substantially increased. The use of an active filler gave better results than inert fillers or fibrous reinforcements. The composition of the joining material, as well as the processing conditions, still needs to be further optimized, in order to increase the shear strength to the required design values.

Acknowledgements

This work was carried out in the frame of the E.C. Fusion Technology Long-Term Programme – Materials 1995/98 SDS 3.1.1.

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