

Characterisation of joined surface modified SiCf/SiC composites

*Original*

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(Article begins on next page)

1 **Characterisation of joined surface modified SiC<sub>f</sub>/SiC composites**

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46  
47 **Abstract**

48  
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50 This work has focused on *surface engineering* coupled with brazing to join SiC<sub>f</sub>/SiC  
51  
52 composites and to improve the joint strength. The surface engineering was carried out  
53  
54 through the Selective Thermal Removal (STR) of SiC fibres from a SiC<sub>f</sub>/SiC composite  
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56 to obtain “brush-like” surfaces; the modified composites were then joined by means of  
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58 an AgCuTi braze alloy. In order to investigate whether a “brush-like” interface could  
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1 enhance the mechanical strength of the joint by increasing mechanical interlocking with  
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3 the brazing alloy at the micron scale, the joints were tested with and without surface  
4  
5 engineering by means of single lap offset under compression.  
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7

8 The treatment to form a roughened surface suitable for mechanical keying of the braze  
9  
10 metal led to an improved ductility of the joint; the fracture surfaces demonstrated that  
11  
12 the proposed method is promising, even though the treatment damages locally the  
13  
14 composite.  
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20 **Keywords:** ceramic matrix composites; joining; surface engineering; brazing.  
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## 25 **1. Introduction**

26  
27 Ceramic Matrix Composites (CMCs) are generally composed of straight or woven  
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29 ceramic fibres embedded in a ceramic matrix with a weak bond between them, a process  
30  
31 that leads to an improved fracture toughness of the composites compared to the matrix  
32  
33 or to fibre materials. Because of their superior thermomechanical properties, CMCs  
34  
35 based on a SiC matrix reinforced with silicon carbide fibres ( $\text{SiC}_f/\text{SiC}$ ) are materials of  
36  
37 great interest for many applications, ranging from the aerospace [1,2] to the nuclear  
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39 field [3-5], which require extremely high temperature stability and damage tolerance.  
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44  $\text{SiC}_f/\text{SiC}$  are currently being used as thermo-structural materials in different  
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46 components, such as aircraft brakes, body flaps, leading edges, heat exchanger  
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48 components, gas turbines for power plants, thermal protection systems for space  
49  
50 vehicles and the inner walls of plasma chambers of nuclear fusion reactors [6].  
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54 However, in many cases, their use depends to a great extent on their ability to be joined  
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56 and integrated, since the manufacturing of these materials as large components or  
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58 complex geometries is difficult and expensive. The study and development of reliable  
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1 joining methods to assemble CMC as large components in complex structures is a  
2  
3 critical issue for a wider use of these materials.  
4

5  
6 The surface engineering of CMC has attracted a great deal of attention from the  
7  
8 scientific community as it represents a smart and still unexplored technology that can be  
9  
10 used to improve the mechanical performances of CMC joints. The design of interfaces  
11  
12 coupled with suitable joining materials and joining technologies are key parameters in  
13  
14 manufacturing, especially for ceramic matrix composite-based components. In a  
15  
16 previous work [7], we proposed a technique based on the Selective Thermal Removal  
17  
18 (STR) of SiC fibres from SiC<sub>f</sub>/SiC composites. This process leads to a micro-sized  
19  
20 “brush-like” structure that is able to enhance the specific surface of SiC<sub>f</sub>/SiC. It was  
21  
22 demonstrated that the specific surface of SiC<sub>f</sub>/SiC significantly increased (~ 90%) when  
23  
24 fibres were removed parallel and perpendicular to the composite surface.  
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28  
29 In [7], the effect of the STR process on the composite surface was discussed from the  
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31 morphological point of view and wetting tests, using a AgCuTi brazing alloy, on as-  
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33 received and modified SiC<sub>f</sub>/SiC were performed. In the present work, starting from  
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35 those promising results, experimental parameters have been transferred from wetting  
36  
37 test to joining process and adhesion issues are presented. Surface modified SiC<sub>f</sub>/SiC  
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39 joined samples have been manufactured and their microstructure studied. In order to  
40  
41 assess the effectiveness of the STR process, mechanical tests (single-lap offset under  
42  
43 compression) have been conducted on STR modified SiC<sub>f</sub>/SiC joined by brazing and the  
44  
45 results have been compared with those of reference SiC<sub>f</sub>/SiC joints (obtained with as-  
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47 received SiC<sub>f</sub>/SiC). Furthermore, the mechanical strength of SiC<sub>f</sub>/SiC after STR has  
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49 been measured to verify the detrimental effects of the surface engineering process on  
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51 the SiC-based composite itself.  
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## 2. Materials and methods

Keraman® SiC<sub>f</sub>/SiC samples supplied by MT Aerospace (Germany) were used for the experimental work. The composites were manufactured at MT Aerospace, Germany, using the standard gradient Chemical Vapour Infiltration (CVI) process and supplied as rectangular shaped samples; they contained Tyranno ® S grade 1.6 K (Ube Industries, Tokyo, Japan) fibres coated with a thin C layer. SiC<sub>f</sub>/SiC has a thermal expansion coefficient of  $4 \times 10^{-6} \text{ K}^{-1}$ .

All the SiC<sub>f</sub>/SiC showed pores between the fabric layers and some cracks on the surface.

Before brazing, the SiC<sub>f</sub>/SiC were cut into 5 mm × 5 mm × 2 mm slabs and were then processed at 1450 °C for 2 hours under flowing Ar in a tubular oven (according to the STR process described in [7]). The brazing alloy used to join the CMC was a commercial AgCuTi alloy (CB4, Degussa, Germany) in the form of 100 μm thick foils. This alloy consists of eutectic Ag-Cu activated with 3 wt. % of Ti (nominal composition: Ag: 57.7, Cu: 36.8, Ti: 5.5 at.%) and belongs to a family of braze media that are commonly used for mid-temperature joining processes; its melting range is 780-805 °C and it has an optimal brazing temperature in the 850-950 °C range. The braze foil was sandwiched between composite blocks and this assembly was introduced into an alumina tubular furnace set at 850 °C, held at this temperature for 30 min and then cooled down at a rate of 5 °C · min<sup>-1</sup>. During the process, a vacuum lower than 5 · 10<sup>-4</sup> Pa was maintained by means of a turbomolecular pump; no external load was necessary as adhesion was assured by the capillary forces of the liquid which pulled the surfaces together.

The SiC<sub>f</sub>/SiC and joined samples were characterised before and after the STR process using Field Emission Scanning Electron Microscopy (FESEM- ZEISS Supra 40) with

1 an Energy Dispersive Spectroscopy (EDS- SW9100 EDAX) detector and scanning  
2  
3 electron microscopy (SEM, model: LEO 1450 VP) with an electron microprobe (EDS)  
4  
5 (Oxford Instruments, 7353 model with Oxford-INCA software v. 4.07, type of detector:  
6  
7 Si(Li)).  
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9

10 Investigation and structural analysis of the interfacial phases were carried out using  
11  
12 Transmission Electron Microscope (TEM) JEOL 2100F UHR operated at 200kV. Image  
13  
14 characterization was done in scanning/transmission mode employing bright field  
15  
16 detector, in order to utilize mass-thickness contrast for enhancing the distinction of  
17  
18 areas with different atomic number and to avoid undesirable effects of dynamical  
19  
20 diffraction. In this mode, EDS analyses were performed by means of detector with  
21  
22 crystal size 80mm<sup>2</sup> (Oxford Instruments, Xmax80, INCA software). Phase identification  
23  
24 was confirmed by Selected Area Electron Diffraction (SAED) technique. For processing  
25  
26 of obtained diffraction patterns licensed softwares were used (Gatan Digital Micrograph  
27  
28 and CMPR database). Calibration of digital camera in reciprocal space was set on  
29  
30 MoO<sub>3</sub> monocrystal standard.  
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37 Thin foils for purposes of TEM observations were prepared using standard preparation  
38  
39 methods including cutting, grinding, polishing, dimpling and final step of the thinning  
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41 procedure was done by ion milling (PIPS Model 691 Gatan) operated at 4.5 kV with  
42  
43 ion-beam angles of 4° and 3°.  
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47 Complementary phases analysis was carried out by X-ray diffraction (XRD, X'Pert Pro  
48  
49 MRD, Panalytical, Cu K<sub>α</sub> radiation, with the aid of the X-Pert High Score software) and  
50  
51 identified with JCPDS files.  
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54 The shear strength of the joined samples (at least 4 as-received SiC<sub>f</sub>/SiC joints and STR  
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56 SiC<sub>f</sub>/SiC joints) was evaluated using the single lap offset (SLO) test under compression  
57  
58 at room temperature, according to a method adapted from ASTM D905-08 (universal  
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1 testing machine SINTEC D/10) [8]; the crosshead speed was 0.5 mm/min. The  
2  
3 maximum force was recorded and the shear strength was calculated by dividing the  
4  
5 maximum force by the joining area.  
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8 The size of the single lap off-set shear tests for brazed samples was 10 mm × 3 mm × 4  
9  
10 mm with a joined area of 30 mm<sup>2</sup>.  
11

12 The SiC<sub>f</sub>/SiC samples were subjected to 3-point bending tests (v= 0.5 mm/min; span=  
13  
14 40 mm), according to ASTM standard C1341-13 [9], at room temperature before and  
15  
16 after the STR treatment. The composite size was 3 mm × 4 mm × 45 mm. The fracture  
17  
18 surfaces of the specimens were investigated after mechanical characterization by means  
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20 of SEM/FESEM.  
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### 30 **3. Results and discussion**

#### 31 **3.1 Surface modification and joining**

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33 The STR process was carried out on the SiC<sub>f</sub>/SiC samples at 1450°C for 2 hours to  
34  
35 obtain “brush-like” surfaces [7]. The appearance of the modified surface is shown in  
36  
37 Figure 1; the SiC fibres were partially and homogeneously removed from the composite  
38  
39 surface, and pores and detached areas formed at the fibre/matrix interface. The presence  
40  
41 of silica microwires is evident on the fibre surface (Figure 1 d) and inset e)), both on the  
42  
43 fibre cross-sectional area (as shown in the micrographs) and on the external part of the  
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45 fibres parallel to the treated surface.  
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51 When exposed to temperatures above 1000°C, the fibres used as reinforcement in the  
52  
53 SiC<sub>f</sub>/SiC are subjected to a thermal degradation, as discussed in [7]. The thermal  
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55 degradation occurs both at the end of the fibres and on the body of the fibres, at least on  
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1 the part of the body exposed to STR atmosphere. In that sense, STR also acts on fibres  
2 parallel to the joining surface and on the overall mechanical properties of the composite.  
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4 The modified SiC<sub>f</sub>/SiC surfaces have been joined using a AgCuTi brazing alloy; the  
5  
6 joining process ,carried out in the present work, is based on the results of the wetting  
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8 tests [7]. The proof of concept of the STR modification surface technique to implement  
9  
10 a joining process has been carried out through the morphological investigation of the  
11  
12 cross-section of the joints; Figures 2 a and b show an example of an SiC<sub>f</sub>/SiC - AgCuTi  
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14 - SiC<sub>f</sub>/SiC joint.  
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20 Both types of composites (as-received (Figure 3) and STR treated (Figure 4) SiC<sub>f</sub>/SiC)  
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22 showed the presence of sound joints with the brazing alloy, which was well distributed  
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24 along the whole joint and had infiltrated into the pores and cracks, as can clearly be  
25  
26 observed in the microstructures in Figures 3 and 4. The microstructures of the joint in  
27  
28 Figure 3 a-c show that a sub-micrometric Ti-rich layer has formed at the interface and it  
29  
30 is visible in the higher magnification figure (inset c). The wetting tests performed by the  
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32 authors [7] showed that the braze perfectly spread on the SiC<sub>f</sub>/SiC surfaces, filling all  
33  
34 the cavities and gaps introduced by STR. The present paper reports on the experimental  
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36 activity carried out to verify the beneficial effect of STR on joints manufacturing. The  
37  
38 thermal treatment etched the fibers not only at their top but also on their side, which in  
39  
40 turn leads to a micro-sized “brush-like” structure on both the facing surfaces of the  
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42 joints.  
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49 An SEM cross section of an STR SiC<sub>f</sub>/SiC joint is shown in Figure 4 for comparison  
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51 purposes: the brazing alloy has infiltrated the composite surface along the macro-  
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53 porosities, in the same way as for the as-received SiC<sub>f</sub>/SiC, but in this case the brazing  
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55 alloy has penetrated for ~ 10 μm and infiltrated the gaps obtained by the partial removal  
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57 of fibres, thus a “brush-like” joint has formed with interlocked joining materials (Figure  
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1 4 b, c). ~~The interfacial composition (see Table 1) was similar for the two types of~~  
2  
3 ~~joints.~~  
4

5 The interfacial composition (see Table 1) was similar for the two types of joints.  
6

7  
8 In order to improve the reliability of data from SEM-EDS, the thin Ti-rich layer was  
9 analysed by TEM to further identify the interfacial products. Figure 5 shows high-  
10 magnification images of the interface and SAED patterns identification of the resulting  
11 phases: going from SiC<sub>f</sub>/SiC towards the brazing alloy, a ~400nm thick layer formed  
12 mostly of TiC nanograins, and a 250 nm thick Ti<sub>5</sub>Si<sub>3</sub> layer were detected.  
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20 In the case of an untreated SiC<sub>f</sub>/SiC surface (Figure 3), the Ti-rich layer is evident along  
21 the whole joint interface and it follows the surface profile (see Figure 3 c). In the case of  
22 an STR SiC<sub>f</sub>/SiC joint (Figure 4), the Ti-rich layer follows the profile of the fibres that  
23 had partially been removed by means of STR (Figure 4 b,c) and the gap between the  
24 fibre and the matrix formed by the thermal selective process. The reaction formed Ti  
25 rich layer was continuous and homogenous, which led to braze wettability on the  
26 surface. This layer followed the profile of the selective etched fibres without obstructing  
27 the cavities, but allowed the molten braze to penetrate. In this way, continuous  
28 infiltration coupled with good wettability were achieved. The formation of this  
29 transition layer at the interface is considered a key factor for ceramic/braze joint  
30 strength. In the case of the SiC<sub>f</sub>/SiC samples, this transition layer was developed on a  
31 larger surface, thus a larger contact area was made available for the adhesion between  
32 the brazing alloy and the SiC<sub>f</sub>/SiC and a gradual CTE (coefficient of thermal expansion)  
33 change took place from the brazing alloy to the composite along the CMC porosities.  
34  
35 The CTEs of the interlayer phases were 18.5 and 8.82 x 10<sup>-6</sup> /°C for the AgCuTi and  
36 Ti<sub>3</sub>SiC<sub>2</sub>, respectively, while the CTE of ~~Ti<sub>5</sub>Si<sub>3</sub>C<sub>x</sub> was assumed equal to that of Ti<sub>5</sub>Si<sub>3</sub>,~~  
37 ~~as the Ti<sub>5</sub>Si<sub>3</sub>C<sub>x</sub> phase is a solid solution that contains only a small percentage of C.~~  
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1 Ti<sub>5</sub>Si<sub>3</sub> exhibits large CTE anisotropy: 6.11 (basal plane) or 16.62 (c-plane) [10]. In  
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3 addition, thermal mismatch may be accommodated for by the relatively good plasticity  
4  
5 of the Ag-Cu braze material [11]. This morphology of the interface in the STR samples  
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7 promotes bonding, provided the interfacial roughness of the treated surface does not  
8  
9 hamper the spread of brazing and physical contact.  
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### 16 **3.2 Mechanical characterization**

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18 The SiC<sub>f</sub>/SiC samples were subjected to 3-point bending tests before and after the STR  
19  
20 treatment in order to assess the possible detrimental effect of the process (1450°C, 2  
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22 hours, Ar atmosphere) on the mechanical properties of the composite. The as-received  
23  
24 SiC<sub>f</sub>/SiC average flexural strength was 576±32 MPa, while the SiC<sub>f</sub>/SiC after STR  
25  
26 showed an average flexural strength of 387±67 MPa.  
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30 Figure 6 shows the representative flexural strength load –displacement curves of the  
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32 SiC<sub>f</sub>/SiC at room temperature, before and after the STR treatment. The STR caused a  
33  
34 decrease in the composites flexural strength of about 30%, due to the fibre and  
35  
36 fibre/matrix interface degradation [7]. The load-displacement curves exhibit the typical  
37  
38 "plastic-like" behaviour that occurs in CMCs because of the toughening mechanisms of  
39  
40 crack bridging, deflection and slippage at the interface between the fibres and matrix;  
41  
42 this trend was observed in both of the tested SiC<sub>f</sub>/SiC types, that is, as-received and STR  
43  
44 treated. The lower mechanical strength of the SiC<sub>f</sub>/SiC after STR may be ascribed to the  
45  
46 partial degradation of the thin carbon layer at the fibre/matrix interface and of the SiC  
47  
48 fibres; it has been reported that the residual tensile strength of Tyranno S fibres, after a  
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50 heat treatment at 1450°C for 1 hour in Ar, is only 20% of the original value [12].  
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57 Despite this significant decrease in mechanical strength of the fibres [13, 14], the  
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59 composite retained about 70% of its original strength after STR. The present  
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1 experimental activity is still underway to confine the STR process to the composite  
2 surface, in order to avoid or significantly reduce the decrease in mechanical strength of  
3 the whole composite.  
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7  
8 The single lap offset (SLO) test under compression was chosen as the mechanical test  
9 for joined SiC<sub>f</sub>/SiC in order to understand whether the STR process was useful to  
10 SiC<sub>f</sub>/SiC to obtain statistically relevant results, and it can be useful to compare different  
11 joining materials and joining processes [15]. The measured shear strength of the joints  
12 produced with the as-received SiC<sub>f</sub>/SiC showed a value of  $21.3 \pm 6.5$  MPa, while the  
13 STR treated and then joined SiC<sub>f</sub>/SiC resulted in a value of  $16.5 \pm 4.5$  MPa, which  
14 roughly corresponds to a 30% decrease, as already observed for the SiC<sub>f</sub>/SiC flexural  
15 strength before and after the same heat treatment used for the STR process.  
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27 The decrease in mechanical strength of the whole SiC<sub>f</sub>/SiC after STR does not allow any  
28 improvement, if any, to be measured in the lap-shear strength of these joints.  
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32 XRD analysis (Figure 7) was carried out on the joint fracture surfaces to detect the  
33 interfacial phases and to confirm the EDS and TEM analysis. The pattern shows several  
34 phases in good agreement with data obtained by TEM analysis: it put in evidence the  
35 formation of Ti<sub>5</sub>Si<sub>3</sub> and TiC (presumably at the interface between the brazing alloy and  
36 the composite surface) and Ag and Cu mostly due the brazing alloy infiltration. The  
37 SiC/SiC composite substrate is detectable by SiC peaks.  
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47 There is no difference in detected phases on the fracture surface after the mechanical  
48 tests on the SiC<sub>f</sub>/SiC joined samples without STR (Figure 7) and with STR.  
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50  
51

52 However, the The fracture surface analysis (Figure 8) revealed a promising  
53 morphology; as can be observed in Figures 8 a) and b), the as-received SiC<sub>f</sub>/SiC joint  
54 underwent delamination of the composite close to the joined region, with a few spots  
55 showing macro-infiltration of the brazing alloy (already shown in the cross-section in  
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1 Figure 3 a) and a flat fracture surface. On the other hand, the infiltration of the brazing  
2 alloy all around the fibres in the STR joined SiC<sub>f</sub>/SiC (Figure 8 d) caused a non-flat  
3 fracture surface and extensive pull-out of the fibres (Figure 8 c), in agreement with the  
4 cross section shown in Figures 4b and c. All this seems to indicate an effective role of  
5 the STR in obtaining a brush-like structure, even though this has not yet been supported  
6 by a measured increase in the mechanical strength of the joint. A schematic of the  
7 fracture path in SiC<sub>f</sub>/SiC joints is shown in Figure 9.

8 The different failure mechanisms in the SiC<sub>f</sub>/SiC joints, with and without the STR  
9 treatment, is also evident in Figure 10. The typical load/displacement curves of the  
10 SiC<sub>f</sub>/SiC joints without STR, as tested by means of SLO, show brittle behaviour (Figure  
11 10, full line). On the other hand, the SiC<sub>f</sub>/SiC processed with STR and then joined  
12 exhibit a load/displacement curve (Figure 10, dotted line) with non-brittle, segmented  
13 behaviour that is typical of a composite material; this can be explained by considering  
14 the increased toughness of the joint, due to the formation of a composite joining  
15 material at the interface between the infiltrated brazing alloys and the STR SiC<sub>f</sub>/SiC.  
16 Other activities are currently underway in order to modify the STR process so as to  
17 avoid the degradation of all the SiC<sub>f</sub>/SiC mechanical properties, that is, to confine the  
18 thermal removal process at the composite surface.

19 Moreover, the stress condition in the braze as in the ceramic composite is clearly  
20 dominated by the residual stresses. Stress level in the braze can be mitigated by plastic  
21 relaxation strain. Contraction of the brazing alloy layer is hindered by the metal  
22 cramped with the ceramic in the structured surface layer.

23 The ceramic opposes the reduction of the distance of the filled holes, thus leading to  
24 compression stresses in the composites. On the other hand, the thermal shrinkage of the  
25 solder in the holes leads to high tension stresses. As a consequence, the brush-like

1 surface can increase the residual stresses at the interface between the metal and the  
2  
3 composite.  
4

5  
6 The tuning of the structuring depth is a key factor in order to have a balance between an  
7  
8 improved mechanical interlocking and a reduction of detrimental stress peaks at the  
9  
10 interface.  
11

#### 12 13 14 15 **4. Conclusions** 16

17  
18 The proposed surface engineering treatment (Selective Thermal Removal, STR) has  
19  
20 been successfully used to obtain “brush-like” SiC<sub>f</sub>/SiC surfaces.  
21

22  
23 The main advantages of this process are (i) modification of the surface, due to a one-  
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25 step pre-treatment, which increases the surface area and allows the joining material to  
26  
27 be mechanically anchored, and (ii) a completely dry and environmental friendly  
28  
29 process, in which no hazardous chemicals are used and which, coupled with a high  
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31 reproducibility of the process, makes this process attractive for implementation in  
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33 automatic and production machinery.  
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37 The brush-like SiC<sub>f</sub>/SiC interface was well infiltrated by the Ag-based brazing alloy,  
38  
39 and this led to a composite-like fracture surface and a “plastic-like” load/displacement  
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41 curve after a single lap offset test under compression of the joints.  
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45 At this stage of the research, the STR process seems to be effective in increasing the  
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47 joint toughness, but the treatment still needs to be modified in order to avoid  
48  
49 degradation of the whole SiC<sub>f</sub>/SiC. The future experimental activities will focus on a  
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51 localised and less harsh thermal process, in order to find a good compromise between  
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53 surface engineering and composite integrity.  
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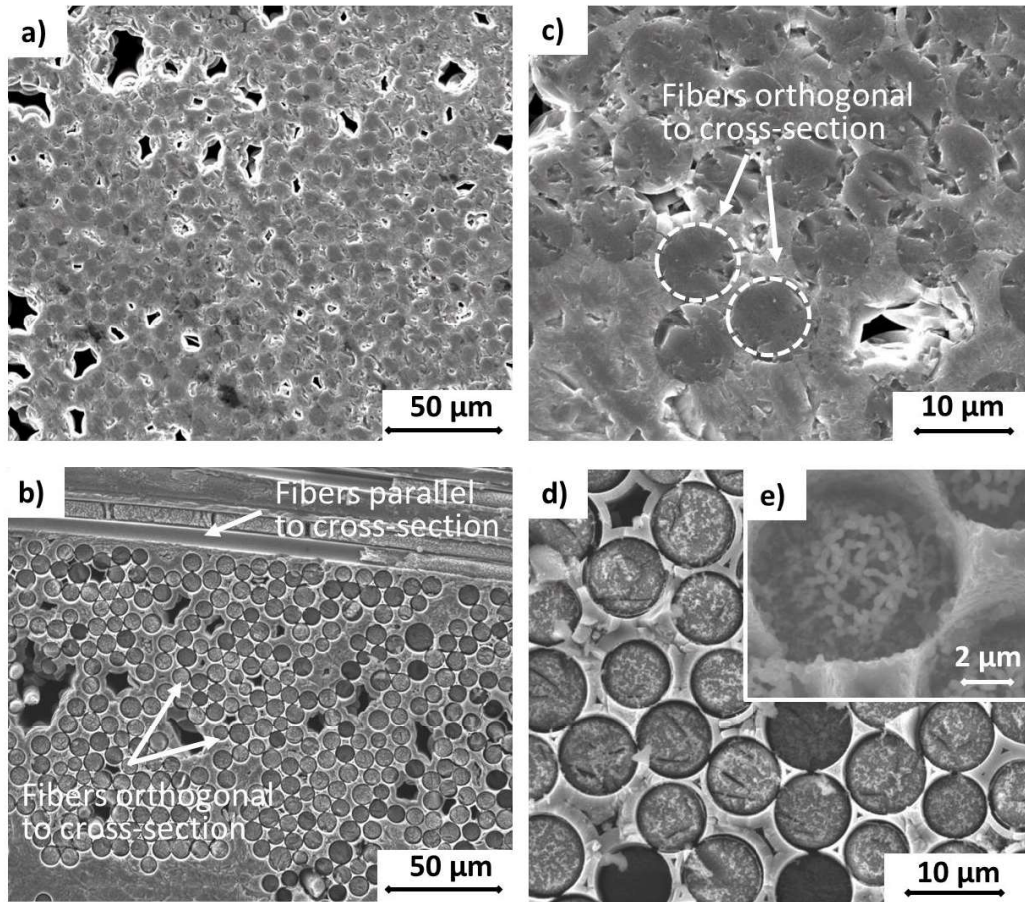
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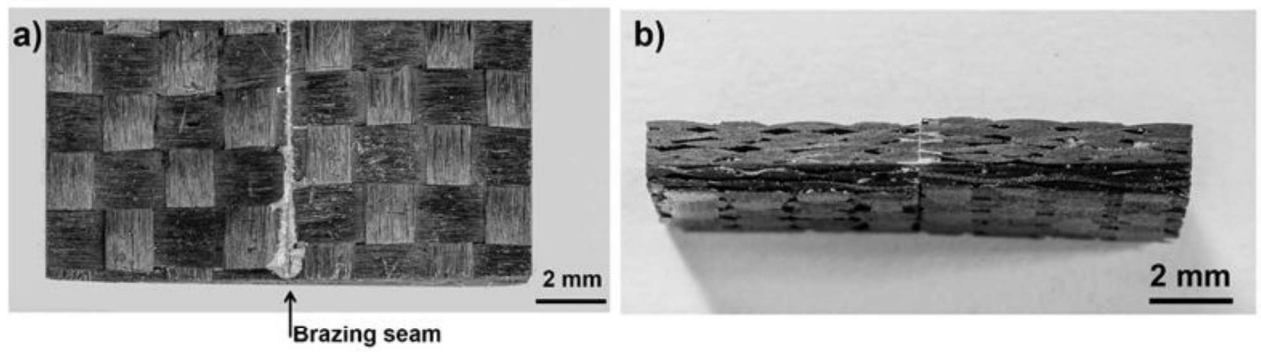


Type of $\text{SiC}_f/\text{SiC}$ composite	Region	Elemental composition (at.%)					Possible phase
		C	Si	Ti	Cu	Ag	
<b>As-received surface (Fig. 3)</b>	A	-	-	0.9	11.3	87.8	Ag(Cu)
	B	-	-	2.8	92.1	5.1	Cu(Ag)
	C (7 kV)	31.5	26.5	35.0	2.7	4.3	$\text{TiC}_x + \text{Ti}_5\text{Si}_3$
<b>STR surface (Fig. 4)</b>	D	-	-	1.1	10.4	88.5	Ag(Cu)
	E	-	-	2.1	93.4	4.5	Cu(Ag)
	F (7 kV)	29.4	24.1	31.5	2.5	3.8	$\text{TiC}_x + \text{Ti}_5\text{Si}_3$

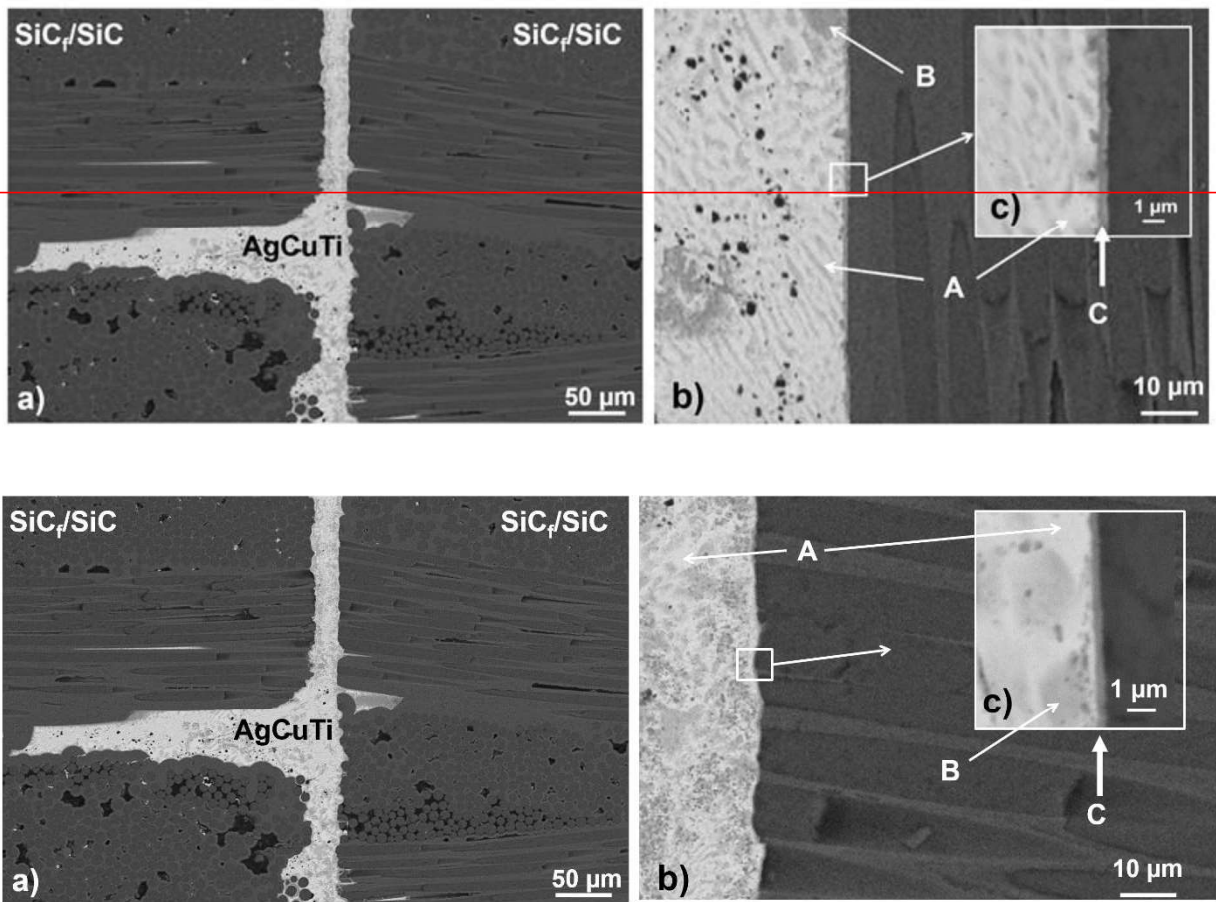
**Table 1:** EDS chemical analysis (at.%) of different regions identified in the cross-sectioned samples; at least 5 spots were analysed for each phase.



**Fig. 1.** SEM-SE images of the surface of the as-received (a,c) and STR treated (b,d)  $\text{SiC}_f/\text{SiC}$ ; inset (e) shows the silica microwires formed on the  $\text{SiC}$  fibre surface.



**Fig. 2.** Macrographs of an SiC<sub>f</sub>/SiC to SiC<sub>f</sub>/SiC joint obtained by means of AgCuTi : the brazing seam is visible (a), as is the infiltration of the braze into the facing composite blocks (b); good adhesion and wetting of the filler alloy along the interface with infiltration of the alloy into the holes and gaps of the composites are visible.



**Fig. 3.** SEM magnification of the cross section of an SiC<sub>f</sub>/SiC joint (as-received composite) showing: a) a panoramic view of the brazing seam showing infiltration into the composite porosities; b) adhesion of the brazing alloy at a microscale and c) higher magnification of the interface.