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Impact of residual stress on thermal damage accumulation, and Young's modulus of fiber-reinforced ultra-high temperature ceramics

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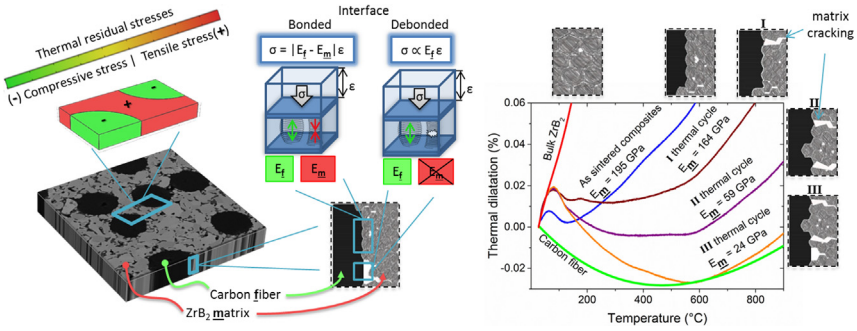
HIGHLIGHTS

- Thermal cycles induce matrix cracking which decreases the Young's modulus of the ZrB_2-based matrix from 195 GPa to 24 GPa.
- The longitudinal thermal dilatation of the composites decreases by thermal cycling and approaches that of bare carbon fibers.
- Contrary to linear elasticity the composite Young's modulus is affected by residual stresses even in absence of microdamage.
- A new equation is proposed to account for residual stresses effect and damage frequency on Young's modulus.

GRAPHICAL ABSTRACT



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ABSTRACT

Ultra-high temperature ceramic matrix composites (*UHTCMCs*) based on ZrB_2-matrix reinforced with 45 vol% of unidirectional continuous carbon fibers are studied through the thermal mechanical hysteresis in order to investigate the thermal damage accumulation. The analysis carried out allowed to extrapolate the Young's modulus of the matrix from thermal expansion measures. It was found that the initial matrix Young's modulus of 195 GPa steadily decreases by thermal cycling the samples between RT and 1300 °C as a consequence of matrix cracking. On the other hand, the analysis suggested that carbon fibers keep their Young's modulus constant at 780 GPa. Finally, the residual stresses due to the different coefficient of thermal expansion between matrix and carbon fibers are discussed and let to justify the Young's modulus of 230 GPa, which cannot be explained with the so-called "rule of mixtures" generally valid and widely used in the composite science.

1. Introduction

The demand for more powerful spacecraft engines and aircrafts with lower emissions is stimulating more research effort to raise the temperature capability of materials well over 1500 °C [1]. In the last decade, the increasing severity of operating conditions of the future hypersonic vehicles [2–4] led to the development of a novel class of ceramic matrix composites (*CMCs*) based on fiber-reinforced ultra-high temperature ceramics (*UHTCs*). Levine et al. in 2002 [5] fabricated a SiC fiber-reinforced ZrB_2 plus 20 vol% SiC composite. However, due to the low fracture toughness, poor resistance to thermal shock, and low resistance to the aggressive oxidation, the authors claimed that the investigated composites were "not ready to be considered as aeropropulsion materials for any applications longer than a few minutes" [5]. The following efforts, on one hand, aimed at enhancing the thermomechanical

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properties/ablation resistance of C/C and C/SiC composites, by adding a certain amount of *UHTC* compounds (minor phases) through powder infiltration (*PI*) [6], chemical vapour infiltration (*CVI*) [7], polymer impregnation pyrolysis (*PIP*) [8], soft-solution [9], slurry infiltration [10], and reactive melt infiltration (*RMI*) [11,12]. On the other hand, continuous fiber-reinforced *UHTCs* produced, since 2004, by *CVI* [13], *RMI* [14], precursor infiltration and pyrolysis process [15] were characterized by wound ceramic matrix, which may strongly affect the material performances. The combination of high flaw tolerance of CMC [16], with the high erosion/oxidation resistance and thermal stability of *UHTCs* (e.g. refractory borides and carbides of early transition metals) [17–20], was achieved by Zoli and Sciti [21] in 2017. They rapidly developed a baseline *UHTCMC* mainly based on ZrB_2 matrix reinforced with around 45 vol% of C_f [22–25]. The flexural strength at RT and 1500 °C, and the fracture toughness were increased up to 360 MPa, 550 MPa, and 11 $MPa \cdot m^{1/2}$, respectively [25,26]. These results, together with the obtained critical flaw size (230–470 μm at RT [25,26], 80 μm at 1500 °C [25]), the retained strength of 310 MPa after water-quenching, from 1500 °C to 20 °C of bath temperature [25], the good oxidation behaviour up to 1500 °C [19], and the absence of appreciable ablation under high velocity oxy-fuel torch (*HVOF*) [2,27,28], confirm the potential of *UHTCMCs* as candidates for reusable thermal protection systems (*TPSs*) for commercial space travel and exploration. In spite of the progress done, *UHTCMCs* still show a considerable margin for further development. In particular, the developed baseline materials may be affected by the coefficient of thermal expansion (*CTE*) mismatch between the matrix and the fiber [20,29]. It was observed that the high *CTE* difference between ZrB_2 and C_f induces flaws formation during cooling from the sintering temperatures, and leaves a high residual stress level that has an impact on the final composite properties [25].

In this work, we investigate the thermomechanical behaviour of unidirectional C_f-reinforced ZrB_2-matrix (the above mentioned "baseline *UHTCMC*") [25] through a simple thermal dilatometry analysis. In spite of its simplicity, our analysis allows to (*i*) characterize the thermal dilatation of the unidirectional UHTCMC along the longitudinal and transverse directions, (*ii*) calculate the anisotropy degree of the composite, and (*iii*) calculate the Young's modulus of the matrix. (*iv*) By thermal cycling the samples, we present a *modus operandi* to characterize the thermal damage accumulation and the thermomechanical stability (a key parameter for the designing of reusable components subjected to thermal cycles). In fact, we show how the evolution of *CTE* response as function of thermal cycles, and the hysteresis loops are correlated to the damage accumulation, and can be used to calculate the Young's modulus variation of the matrix. Finally, by taking into account the residual stresses, a novel model for the linear elastic zone is presented and discussed. This model is meant to replace the first zone of the *ACK* model [30] where the slope of the stress-strain curve should follow the rule of mixtures of the main constituents. This last achievement can bring enormous benefits to the advanced modeling and design of *CMC*-components mechanical behaviour, which is still an open issue [31,32]. In particular, the proposed model (*i*) allows a better prediction and understanding of the elastic properties and progressive damage of different composite materials, and (*ii*) will push forward the application of *CMC*-components in the fields of aerospace and civil engineering.

2. Material and methods

UHTCMCs based on carbon fiber-reinforced ZrB_2 were produced by hot pressing at 1900 °C. The final microstructure consists in 55 vol% of matrix (83% ZrB_2 + 10% SiC + 7% pores), and 45 vol% of homogeneously distributed unidirectional carbon fiber, C_f (so forth called *UD*). Further details on slurry preparation, infiltration, densification, and microstructural and mechanical features of the produced samples are reported in previously published work [25].

The microstructure was analyzed on polished and fractured surfaces by field emission scanning electron microscopy (FE-SEM, Carl Zeiss Sigma NTS Gmbh Öberkochen, Germany).

The relative dimensional change ($\Delta L/L_0$) versus temperature was measured up to 1300 °C under flowing argon with a 5 °C/min heating rate using a dilatometer Netzsch mod. DIL E 402 (Netzsch, Geraetebau, Germany), on 25 mm × 2.5 mm × 2 mm bars. $\Delta L/L_0$ of *UD* was measured along transverse and longitudinal orientation of the C_f. The thermal cycling up to 1300 °C was carried out on the longitudinal direction of C_f, where the thermal dilatations mismatch between matrix and fiber is maximized.

3. Results and discussion

3.1. Microstructural features

The typical microstructure of the baseline composite (*UD*) is shown in Fig. 1. The fibers are all aligned in the same direction in order to create an simplified texture for analysis and interpretation of the mechanical behaviour. The transverse cross section (Fig. 1(a)) displays a homogeneous fiber distribution in the dense matrix and the jagged fiber/matrix interfaces. The longitudinal cross sections show the anisotropic microstructure of the carbon fibers (Fig. 1(b)), the presence of voids between the graphitic layers (Fig. 1(b)), the interlocked fiber/matrix interfaces (Fig. 1(b)), and the cracks in the matrix (Fig. 1(c)) that are homogeneously spaced with a periodicity of 20–40 μm.

3.2. Thermomechanical analysis: a powerful method to determine CTE, coefficient of anisotropy, matrix modulus

Experimental curves in Fig. 2(a) show the relative dimensional change ($\Delta L/L_0$) versus temperature up to 1500 °C for polycrystalline ZrB_2 [20], typical anisotropic pitch-derived carbon fibers (C_f) [29], and unidirectional C_f-reinforced ZrB_2-matrix of this work (*UD* samples), in both directions: along the transverse (*t* subscript) and longitudinal direction (*l* subscript). It can be seen that for C_f, $\Delta L/L_0$ along the transverse section is strongly dependent on the temperature, but nearly constant in the longitudinal direction. Also, we can observe that the elongations ($\Delta L/L_0$) in the transverse direction of the *UD* composite and of C_f are very close to each other and also very comparable with that of a bulk isotropic ZrB_2. This is consistent with the microstructural observation that few or no microcracks were generally found in the ZrB_2 matrix in the transverse cross section of the composite. Looking more carefully, we can see that C_f displays a slightly higher value within the entire investigated temperature range than the isotropic ZrB_2 material.

The higher *CTE* of C_f and the presence of the weak van der Waals interactions between the carbon layers within the fibers could be the reason of the intensification of voids observed in the C_f sections after densification, with respect to those occasionally present in the as-produced fibers [33]. During cooling, void formation compensates the higher transverse thermal shrinkage of C_f. In fact, since C_f are physically well-joined with the ZrB_2 matrix – it was found out that matrix shrinkage during the densification process produces strongly jagged ZrB_2/C_f [26] – the biaxial tension (along the transverse directions) and uniaxial compression (along the longitudinal direction), seems to favour the carbon layers sliding among themselves rather than delamination at the ZrB_2/C_f interface. This mechanical interlocking developed at the ZrB_2/C_f interfaces transfers the C_f anisotropy to the *UD* samples: for this reason $(\Delta L/L_0)_{UD,l}$ is much lower and much less dependent on the temperature compared to $(\Delta L/L_0)_{UD,t}$ (Fig. 2(a)). This aspect should be taken into account especially when *UHCMCs* are designed with more complex structure with respect to the unidirectional configuration, e.g. 0°/90° cross-ply configuration.

Let's us now evaluate the degree of anisotropy as function of temperature, $R(T)$ of *UD* composite and C_f. For the composite and the fiber, R is calculated using the experimental values of transverse (*t*) and the

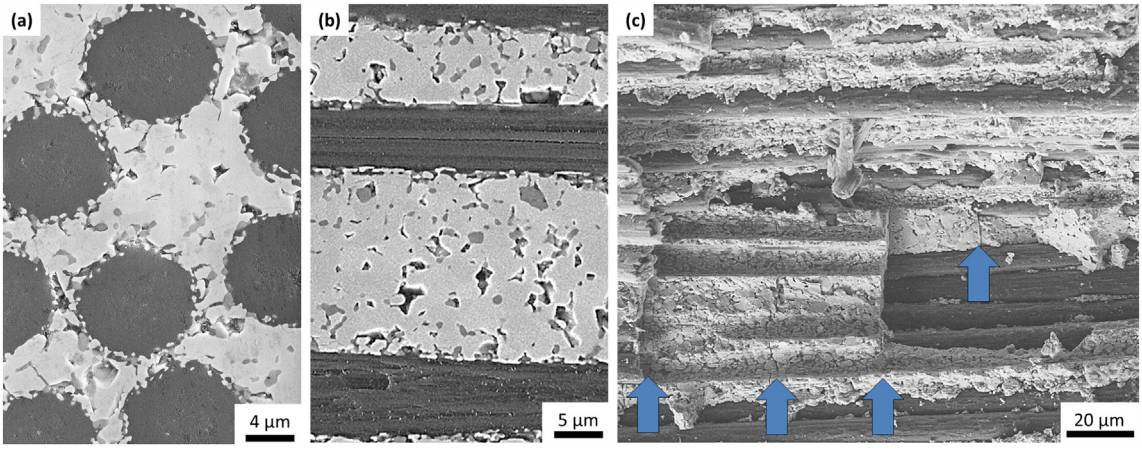

Fig. 1. SEM micrographs of polished cross section of *UHTCMC* based on pitch-derived carbon fibers and ZrB_2/SiC matrix along (a) the transverse direction, and (b) longitudinal direction. (c) Fracture surface along the longitudinal direction (the arrows point the presence of cracks in the matrix).

longitudinal (l) relative dimensional change ($\Delta L/L_0$) [29]:

$$R(T) = \frac{\left(1 + \left(\frac{\Delta L}{L_0}\right)_t\right)}{\left(1 + \left(\frac{\Delta L}{L_0}\right)_l\right)} - 1 \quad (1)$$

For instance, in the case of isotropic bulk ZrB_2 $R(T) = 0$. The values obtained from Eq. (1) are plotted in Fig. 2(b). On the other hand, for the sake of comparison, we calculated the theoretical value of $R(T)$ for the *UD* composite with the micromechanical model, making the following assumptions:

- the two components are perfectly bonded together forming a simple beam so that they deform together (i.e. iso-strain condition);
- no external load is applied so that the overall stresses in the longitudinal direction of the beam are balanced (i.e. equilibrium condition);
- along the transverse direction there is no *CTE* difference between the matrix and the transversely isotropic fibers [33]. Hence the composites *CTE* is the same as the side phases.

In the above conditions, the theoretical $(\Delta L/L_0)_{l,c}$ of the *UD* sample is modeled by the Schapery's model [34]:

$$\left(\frac{\Delta L}{L_0}\right)_{l,c} = \frac{\left(\frac{\Delta L}{L_0}\right)_{l,f} E_f V_f + \left(\frac{\Delta L}{L_0}\right)_m E_m (1 - V_f)}{E_f V_f + E_m (1 - V_f)} \quad (2)$$

where the subscripts 'm', 'f', and 'c' refer to matrix, fiber, and composite, respectively, V is the volume fraction, and E is the Young's modulus. In our case, the theoretical $(\Delta L/L_0)_{l,c}$ of the *UD* sample is calculated by fixing $V_f = 0.45$, $E_f = 780$ GPa, and $E_m = 420$ GPa [20], and imposing thermal dilatation of the side phases: C_f [29], and ZrB_2 [20]. Hence, the theoretical $R(T)$ is obtained by using in the denominator of Eq. (1) the theoretical $(\Delta L/L_0)_{l,c}$ instead of the experimental value ($(\Delta L/L_0)_{UD,l}$ curve plotted in Fig. 2(a)). It is worthy to note that, along the transverse direction, there is not significant difference between the theoretical $(\Delta L/L_0)_{t,c}$ and the experimental one, since $(\Delta L/L_0)_{Cf,t}$ and $(\Delta L/L_0)_{ZrB2,t}$ are quite similar to $(\Delta L/L_0)_{UD,t}$. Looking at the curves in Fig. 2(b), we can see that the anisotropy, $R(T)$, increases with increasing the temperature for C_f as well as for *UD* samples. At 1500 °C the anisotropic degree value for C_f is 1.03, whilst in the *UD* composite R is reduced to 0.71 owing to the isotropic behaviour of the matrix ($R = 0$). However, the obtained value of $R(1500\ °C) = 0.71$ is much higher than the expected one: $R(1500\ °C) = 0.39$. The key to understand the reasons for this discrepancy is in the longitudinal $\Delta L/L_0$ contribution of the *UD* sample, that should be much higher.

In fact, as it can be seen in Fig. 3, the theoretical $(\Delta L/L_0)_l$, calculated trough the Eq. (2), is higher than the experimental one. This is usually due to mechanical discontinuities in the materials such as pores and/or cracks. Since the voids inside C_f do not affect the mechanical behaviour in the longitudinal direction, we believe that a major role is played by matrix cracking [33]. Cracks have a strong detrimental effect on the

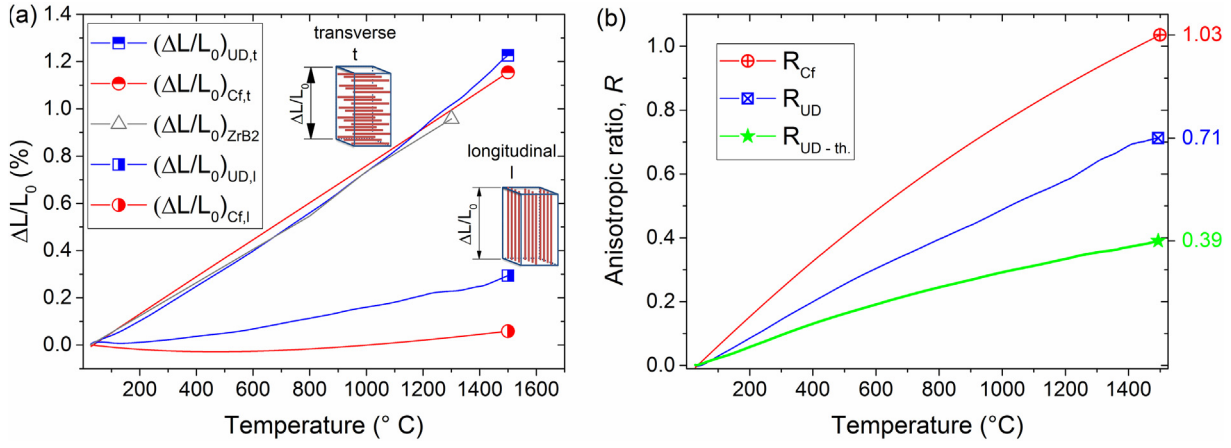

Fig. 2. (a) Relative dimensional change ($\Delta L/L_0$) vs. temperature of the tested *UD* samples (blue line with square symbols), and pitch-derived fibers C_f [29] (red line with circle symbols) along the transverse (t subscript, half up symbols) and the longitudinal (l subscript, half right symbols) direction, and polycrystalline ZrB_2 [20] (gray line with open triangle symbol). (b) Anisotropy degree R vs. temperature of tested *UD* (blue line with marked square symbol), C_f (red line with marked circle symbol), and theoretical *UD* trend (green line with solid star symbol).

E_m value owing to their high surface/volume ratio. This detrimental effect is commonly modeled by the following exponential relationship [35]:

$$E = E_{th}e^{-bp} \tag{3}$$

where E, and E_{th} are the Young's modulus of the real material and the fully dense one, respectively, p is the voids volumetric fraction, and b is a numerical constant that takes into account the shape and stacking of the voids. The value of the latter is attested in the range of about 2–4 in case of sintering porosity, and can be higher than 9 in the stricter conditions as in the presence of multiple interacting cracks [36]. Thus, considering $b = 9$–13, $p = 5$–10 vol%, and $E_{th} = 420$ GPa, the Young's modulus of the matrix could be attested between 114 GPa and 268 GPa. A measure of E_m can be extrapolated by fitting the experimental values ($(\Delta L/L_0)_{UD,l}$) with Eq. (2). By releasing only the E_m parameter into the theoretical $(\Delta L/L_0)_{l,c}$ (Eq. (2)), the fitting convergence is reached by decreasing E_m from 420 GPa down to 195 GPa (Fig. 3). Similarly, the matrix cracking causes the much lower flexural strength, $\sigma =$ 65 MPa [25], of the matrix with respect to the corresponding bulk material $\sigma = 600$ MPa [20]. All the above considerations suggest that mechanical and thermomechanical behaviours of the *UD* samples are mainly controlled by C_f simply because the matrix is strongly damaged.

3.3. Residual stress build up and matrix cracking

Cracks in the matrix are generated by residual stresses developed during cooling in case that neither the matrix nor the fibers exhibit time-dependent deformation (i.e. creep). In fact, during cooling under the 'joining' temperature of about 1550 °C [36] the internal stresses, σ_0, should follow the trends plotted in Fig. 4(b), which were calculated with the following equations:

$$\sigma_{0,f} = \left[\left(\frac{\Delta L}{L_0}\right)_{l,f} - \left(\frac{\Delta L}{L_0}\right)_{l,c}\right] E_f \tag{4}$$

$$\sigma_{0,m} = \left[\left(\frac{\Delta L}{L_0}\right)_{m} - \left(\frac{\Delta L}{L_0}\right)_{l,c}\right] E_m \tag{5}$$

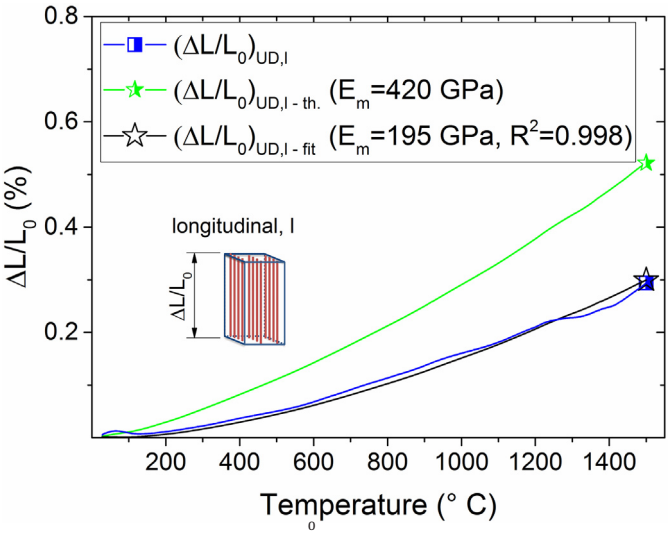

Fig. 3. Relative dimensional change ($\Delta L/L$) along *l*-direction vs. temperature of the tested *UD* samples (blue line with square symbols), theoretical trend (green line with half right star symbols), and adjusted trend which take into account the matrix damaging/softening (black line with open star symbols).

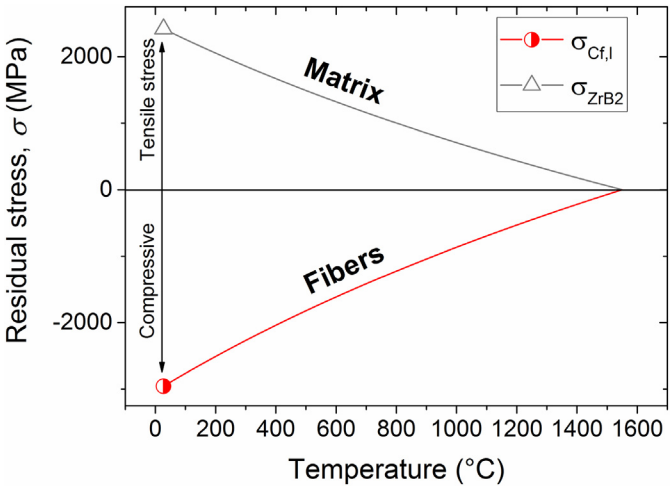

Fig. 4. Theoretical residual stresses into matrix (gray line with open triangle symbol) and fibers (red line with half right circle symbols) that should develop during cooling.

In the above equations the Young's moduli are considered temperature dependent in order to account for the material softening with the temperature. In particular E_m is set to follow the empirical Watchman's equation [37,38]:

$$E_m(T) = E_m - bT\, e^{\frac{-T_0}{T}} \tag{6}$$

where the constants b and T_0 are fixed to 0.0399 and 193 K [38]. The magnitude of the calculated stresses is lower than the tensile strength of the fibers (which is over 4 GPa), but is not small enough to avoid the local failure of the matrix. According to Eq. (5) a thermal difference of 365 °C is enough to overcome the tensile strength of the ZrB_2 and should lead to cracks formation in order to allow the matrix shrinkage which is accompanied by stress release. Unfortunately, it is not easy to quantify the residual stress in the matrix, and to understand the effect of the high level of compressional stress into the fibers. All the performed microstructure analyses on the produced *UHTCMCs* shown no evidence of macroscopic fiber kink or buckling. This lack of evidence is reasonable due to the high stiffness of the matrix, but, inside the fibers, kink and buckling phenomena could occur at the scale of the alignment of graphene sheets along the fiber axis. The SEM analysis confirms the presence of the pores (Fig. 1(b)) and cracks (Fig. 1(c)) due to the biaxial and uniaxial tension, respectively, developed during cooling. In particular, it is worthy to note how the biaxial tension not only produces pores inside C_f, but, in some case, where C_f are not well distributed, we observe a strong deformation of C_f shape (not shown here). On the other side, along the longitudinal direction, it is not surprising that the high *CTE* difference leads to a small crack spacing which is attested at about 20–40 μm (Fig. 1(c)). The evidence of the transverse cracks is in accordance with the above calculated build up in residual stress (1.6 MPa/°C of average value of the residual longitudinal stress) and with the fact that the load between matrix and fibers is transmitted by shear stresses which cannot rise above the matrix shear strength.

3.4. Thermal cycling and damage accumulation

Owing to the high *CTE* difference between matrix and fibers, it is reasonable to suppose that the observed crack spacing of 20–40 μm (as point out by the arrows in Fig. 1(c)) is not that of equilibrium. In fact, by thermomechanical cycling the *UD* samples, hystereses were found (Fig. 5(a)), and should be mainly attributed to the crack formation. Moreover, the cracks formation at each thermal cycle leads to a residual permanent expansion which is the consequence of the compressional

strains released by fibers and the discontinuity produced in the matrix. It can be seen that the initial dimension at *RT*, L_0, gradually increases up to 0.11%, 0.19%, and 0.35% after the first, second, and third thermal cycle, respectively. By rescaling the measured $(\Delta L/L_0)$ curves in agreement with these increments (Fig. 5(b)), it can be seen how the $(\Delta L/L_0)$ curves gradually approach the $(\frac{\Delta L}{L_0})_{l,f}$ trend. This suggests that the fibers keep their initial modulus, which, becoming increasingly bigger than matrix modulus, imposes their thermal expansion [39]. In particular, with the third rising (III-5 curve) at about 550 °C–750 °C, *UD* sample expands as free C_f. This means that cracks formed during thermal cycles allow the free expansion and shrinkage of the matrix and fibers, respectively, up to 750 °C, roughly. The free matrix expansion is highlighted by the presence of the "bell" between *RT* and 100 °C, and the free fiber shrinkage is clearly visible by the $(\Delta L/L_0)$ decreasing up to 600 °C. Above 600 °C the higher matrix *CTE* pushes up the $(\Delta L/L_0)$ values of the *UD* sample with respect to that of C_f. In this way, stresses start to increase again and can produce further cracks formation as it is evidenced by the slope increase above 1200 °C, which produces the matrix expansion (detected by the thermomechanical analysis, *TMA*) and the fiber shrinkage (not detectable). The decline of the elastic properties of the matrix was quantified by fitting $(\Delta L/L_0)$ curves with Eq. (2). The fitting (Fig. 5(b)) was performed only in the part of the $(\Delta L/L_0)$ curves where matrix and fiber mechanically interact (i.e. mechanical constrained) without producing evident cracks formation. This reasonably occurs in the range between 650 °C and 1000 °C. The fitting (Fig. 5(b)) shows a steady decrease of E_m which goes from the 195 GPa of the as sintered sample to 164 GPa with the first thermal cycle and then drops down to 59 GPa and 24 GPa with the following 2 cycles. This means that the high *CTE* difference severely afflicts the matrix which from one side releases its residual tensile stress, but on the other losses its mechanical continuity [40–42]. Anyway, the residual stress release by the cracks formation could be a minor concern, since these materials are intrinsic 'damage-tolerant' and 'notch-insensitive', show an increase of strength after thermal shock, and can be shocked at temperatures higher than 1480 °C [25]. This behaviour has been correlated to cracks formation under the critical value per unit volume after thermal shock which has led to the residual stress release. Similar results were obtained with fatigue tests where the fully-developed micro-damaging led to more uniform stress state and a consecutive residual strength higher than the pristine ultimate strength [42,43]. On the other hand, the fiber compression state can be exploited for structural application at RT, since the enhanced crack bridging by the fiber – greater applied stress must be imposed on the fibers to attain their fracture stresses - enhances the toughening [44].

3.5. Composite Young's modulus and the "springs' theory"

The effect of residual stresses - caused by the matrix/fiber CTE mismatch when cooling to ambient temperature after densification – on mechanical strength at RT and high temperature, fatigue damage, fatigue life, and thermal shock was studied by several authors [25,41,42,45]. However, no explanation, to the best of our knowledge, is given about the effect of the residual stresses on the macroscopic elastic modulus prior to microdamage development. In ref. [45], Ziegler shows how the thermal stresses influence the flexural strength of SiC fiber-reinforced glass composites, but no attempts are made to explain the Young modulus decrease from about 125 GPa to 75 GPa with the increasing of the thermal stress from $2.09 \cdot 10^{-6}\,°C^{-1}$ to $6.6 \cdot 10^{-6}\,°C^{-1}$. Reynaud et al. [46] show that tensile mean elastic modulus of SiC/SiC *CMC* reaches the value of the compressive mean elastic modulus when thermal residual stresses are released under fatigue test, but they did not provide a suitable explanation. Shi et al. [47] demonstrate how the residual stress influence the effective Young's modulus of cantilever beams. In our case, since the residual stresses should be close to the critical values, the Young's modulus should be strongly affected. In fact, in our previous work [25], we found out a longitudinal Young's modulus of 232 ± 10 GPa which is completely in disagreement with the expected one calculated with the following Voigt's equation (the conventional rule of mixture):

$$E_c = E_f V_f + E_m(1-V_f) \tag{7}$$

Whatever value from 420 down to 0 GPa is attributed to E_m, Eq. (6) gives a higher result with respect to the value experimentally measured. This simple observation demonstrates that the decrease in the Young modulus of *UHTCMCs* cannot be attributed just to matrix damage [48], but a direct contribution from residual stresses should be added even in absence of defects. In fact, since the fibers are compressed and the matrix is stretched, they would tend to expand and shrink, respectively. Hence, when an external tensile force is applied the matrix will counteract the applied tensile force as prescribed by Hooke's law, and the fibers (in the isostrain condition: $\varepsilon_m = \varepsilon_f = \varepsilon$) will spontaneously follow the expansion releasing part of the bounded energy due to the initial thermal stress ($\sigma_{0,f}$) and strain ($\varepsilon_{0,f}$): $U_{0,f} = \frac{1}{2}\int_V (\sigma_{0,f}\varepsilon_{0,f})dV$. In the opposite case, under external compressional force, the compressed fibers will promptly respond to the mechanical stimulus by increasing their compressional strain according to their stiffness ($E_f = 780$ GPa since the anisotropic pith-derived fiber are just in an initial compressional state and are constrained with the hard matrix, the compressional modulus value

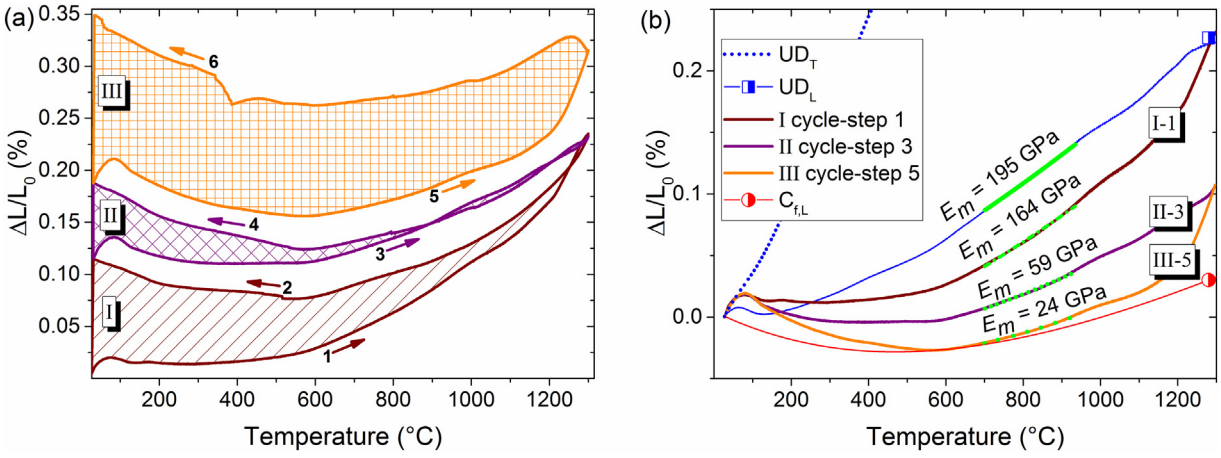

Fig. 5. (a) Relative dimensional change $(\Delta L/L_0)$ along *l*-direction during three consecutive thermal cycles (labeled with roman numbers) between *RT* and 1300 °C. Each thermal step is numbered in chronological order. (b) Relative dimensional change $(\Delta L/L_0)$ along *l*-direction for each heating step together with the $\Delta L/L_0$ of typical anisotropic pitch-derived carbon fibers (C_f) [27], and that of *UD* samples after sintering (plotted also in Fig. 2). The curves were scaled by updating the L_0 value after each thermal cycle.

can be considered equal to the tensile one) whilst the matrix will supply the bounded energy: $U_{0,m} = \frac{1}{2}\int_V (\sigma_{0,m}\varepsilon_{0,m})dV$. The above discussion can be visualized in the sketch of Fig. 6.

According to the linear theory of elasticity, when an external force (F) is applied the variation in strain energy density ($W = U/dV$) is:

$$\Delta W = \left(\frac{1}{2}(\sigma_{0,f} + E_f\varepsilon)(\varepsilon_{0,f} + \varepsilon) + \frac{1}{2}(\sigma_{0,m} + E_m\varepsilon)(\varepsilon_{0,m} + \varepsilon)\right) - (W_{0,f} + W_{0,m}) \tag{8}$$

Since the residuals stresses ($\sigma_{0,m}$ and $\sigma_{0,f}$) are equilibrated in absence of external forces, and the directions of strain and stress are the same in one case and opposite in the other case, Eq. (8) reduces to:

$$\Delta W = \frac{1}{2}|E_f - E_m|\varepsilon^2 \tag{9}$$

The variation in stress, due to the variation in strain imposed by the external force, is:

$$\frac{d\Delta W}{d\varepsilon} = d\sigma = |E_f - E_m|d\varepsilon \tag{10}$$

Hence, the equivalent longitudinal elastic modulus of a portion of unidirectional continuous fiber-reinforced composite, where matrix/fiber interface are not debonded and the entire matrix and fiber volumes are strained, is:

$$E_{eq} = |E_f V_f - E_m(1 - V_f)| \tag{11}$$

Assuming that close to the matrix/fiber interface matrix cracks and to the debonded area has no longer residual stress, the overall elastic modulus of the fiber can be described as a series of thermally strained and unstrained fibers. Thus, the overall fiber elastic modulus can be modeled through the Reuss' approximation (the as called inverse rule of mixture):

$$E_{eq,f} = \frac{E_{eq}\,E_f}{E_f d + E_{eq}(1-d)} \tag{12}$$

where d is fraction of debonded fiber/matrix interface. Since the observed crack frequency is 0.05–0.1 μm^{-1}, the overall Young's modulus calculated trough the mixed rule between the Eqs. (3) and (12) is 222 ± 18 GPa, which matches the measured value.

It is evident that the high matrix/fiber *CTE* mismatch, the jagged matrix/fiber interface (which requires a change in the concept of pure shear load transfer [49,50]), and other strong chemo-physical differences between matrix and fiber, configure the *UHTCMCs* as a novel class of composites materials. Although this class of materials presents critical issues, it has also shown superior properties for the aerospace applications and more generally harsh environments. Hence, in our opinion it is important to dedicate much more effort in the study and development of *UHTCMCs* in order to exploit their potential as much as possible.

4. Conclusion

The present work has shown that (*i*) the strong C_f/ZrB_2 interfaces transfer the anisotropic behaviour of C_f to the entire composite. (*ii*) The high *CTE* mismatch between ZrB_2 based matrix and pitch-derived carbon fiber along their longitudinal direction causes cracks formation during cooling from the sintering temperature. (*iii*) A general modus operandi based on the Schapery's model was suggested to estimate the Young's modulus of the single components, i.e. matrix and fiber. (*iv*) The as sintered matrix is characterized by a Young's modulus of 195 GPa, instead of 420 GPa shown by similar bulk ceramics, due to the cracks formation. (*v*) The longitudinal thermal dilatation decreases by thermal cycling and, due to the damage accumulation, gradually approaches the dilatation behaviour of bare carbon fibers. (*vi*) The experimental data in agreement with the Schapery's model suggest that carbon fibers are affected by neither sintering process nor thermal cycles in Ar atmosphere. (*vii*) Taking into account the damaging of the matrix and the residual stresses distribution, we proposed a modification - called the “springs' theory” - of the law of mixtures in order to explain the measured 230 GPa of Young's modulus for the *UD* composites along the longitudinal direction. These results are particularly important for the understanding of this novel class of materials and paves the way for new developments of different composite materials. On one side, the proposed method to calculate the Young's modulus of the matrix is useful to easily check the matrix contribution to the overall composite behaviour and its stability to the thermal cycles. On the other side, the proposed model is of great value for the accurate prediction and assessment of the elastic and failure properties of *CMC*-components.

Credit authorship contribution statement

L.Z. developed the process of UHTCMCs manufacturing and provided the studied material. **P.G.** formulated the overarching research aims, developed the methodology, created the models, developed the formal analysis, prepared the visualization/data presentation, wrote the manuscript. **D.S.** had the leadership, management and coordination responsibility for the research activity planning and execution, acquired

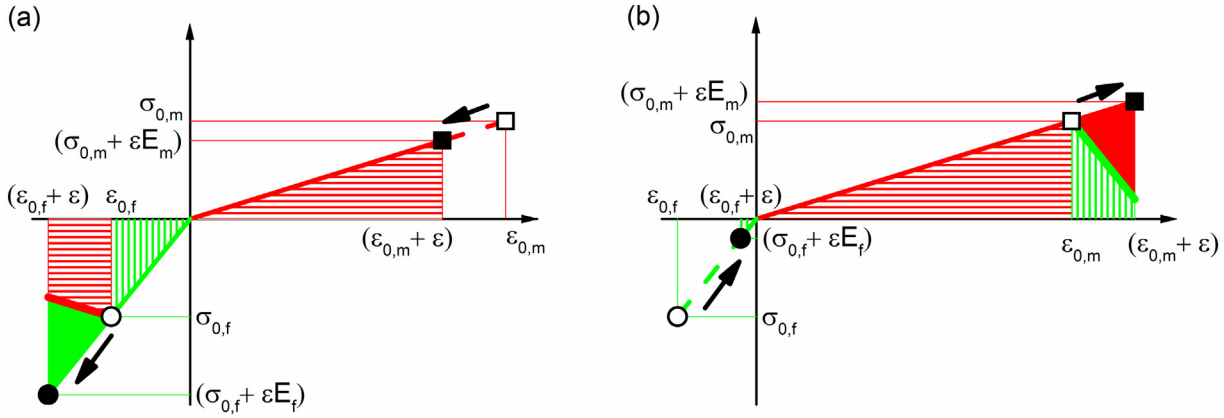

Fig. 6. Scheme of the stress – strain variation after applying a compressive (a) and tensile (b) external load into a continuous fiber-reinforced composite with initial/residual thermal stresses. The arrows flow from the initial point (open symbols) to the final point (solid symbols) for both fiber (circle symbols) and matrix (square symbols). The areas filled with lines pattern (horizontal for matrix, vertical for fiber) correspond to the initial/bounded energy density. The solid areas correspond to deformation work done by the external applied force.

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