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SiC-Matrix Composites: Nonbrittle Ceramics for Thermo-Structural Application

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C/SiC and SiC/SiC composites are tough ceramics when the fiber–matrix bonding is properly optimized, usually through a thin layer of an interfacial material referred to as the interphase. These composites can be fabricated by a variety of techniques that are briefly described and compared. The design of the interphase, matrix, and coating at the nanometer scale, in order to promote microcrack deflection and to enhance the oxidation resistance is discussed. Selected properties of the composites are presented and discussed. Examples of application in engines, heat shields, braking systems, and high-temperature nuclear reactors are shown to illustrate the potential of these materials and the key points that still require research and development.

Introduction

The ceramic matrix composites (CMCs) considered here consist of ceramic fibers (mainly carbon- or SiC-based fibers, generally arranged in multidirectional preforms) embedded in a SiC-matrix. They have been first imagined to replace the carbon/carbon (C/C) composites in long-term application at high temperature (HT) when the atmosphere is oxidizing.[1,2] It was further discovered in a fortuitous manner, and then confirmed by theoretical considerations, that CMCs could display a nonbrittle behavior if the fiber–matrix (FM) bonding was lowered enough, e.g., through the *in situ* formation of a suitable interfacial reaction zone[3] or the use of a so-called *interphase* deposited on the fiber before the infiltration of the matrix.[4,5] The damage-tolerant character of these CMCs, which is an outstanding property for a

ceramic material, potentially allows their use as structural materials for HT application in corrosive atmospheres and explains the tremendous effort of research and development in this field.

Although SiC-matrix composites are promising materials for application in severe environments, their development raises a number of issues that will be discussed in terms of processing, material design, main properties, and actual or potential applications in various fields. A more detailed analysis has been recently presented elsewhere.[6]

Processing

CMCs are fabricated according to gas phase routes (CVI: chemical vapor infiltration), liquid phase routes either from polymers (PIP: polymer impregnation and pyrolysis), or molten elements reacting with the preforms or the atmosphere (RMI: reactive melt

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infiltration), or finally the so-called ceramic or slurry routes (SI–HP: slurry infiltration and hot processing), each displaying advantages and drawbacks. Generally speaking, the matrix should be homogeneously distributed in the preform with limited residual porosity and the FM-bonding well controlled with no significant fiber degradation. Further, the process should be flexible with limited handling and yield near net shape composites, in order to lower production cost.

In the CVI-process, the interphase, the matrix, and the seal-coating (used to seal the open residual porosity and enhance the oxidation resistance) are successively deposited from gaseous precursors. In conventional CVI (referred to as I-CVI, I standing for isothermal/isobaric), there are no temperature/pressure gradients in the fiber preform.[1,2,7] I-CVI is a low-temperature (typically, 900–1100°C), low-pressure (<100 kPa) process, yielding near net shape composites with limited fiber degradation and materials of high microstructural quality. It is also a highly flexible process, a large number of preforms (which could be different in size and shapes) being treated simultaneously with limited handling, in large infiltration furnaces. All these features justify that I-CVI has been rapidly transferred from the laboratory to the plant levels. Conversely, in I-CVI, the densification rate is relatively slow and the residual porosity is significant (typically, 10–15%). The densification rate can be actually improved by applying to the preform a temperature gradient (TG-CVI), a pressure gradient (P-CVI), or both (as in forced or F-CVI), but it is at the expense of flexibility (some fixturing being necessary for each preform to create the gradient(s).[8] It can also be improved by performing intermediate surface machining (to re-open the porosity) but that requires additional handling and raises the fabrication cost. Residual porosity (which is detrimental to thermal conductivity and oxidation resistance) is usually sealed by depositing on the external surface of the composites a suitable coating at the end of the process.

In the RMI (or more simply, MI) process, the fiber preform is first consolidated with carbon (deposited on the coated fibers, e.g., by PIP) and then impregnated with liquid silicon (or an Si alloy), silicon reacting exothermally with the carbon to form *in situ* the SiC-based matrix. RMI is a HT process (1400–1600°C) and liquid silicon is a highly reactive medium. Hence, it can be used only with fibers of high thermal stability (carbon or oxygen-free SiC-based fibers) protected with a suitable interphase, e.g., dual pyrocarbon/SiC or boron nitride (BN)/SiC interphases where the SiC-sublayer acts as a diffusion barrier.[9] Further, the matrix is rarely pure SiC but a mixture of SiC and free silicon (free silicon lowering its refractoriness and creep resistance), however, the content of the latter can be limited if liquid silicon is replaced by a suitable silicon alloy. On the other hand, RMI is a fast densification technique and the corresponding composites are near net shape with low residual porosity ($V_p<5\%$).

I-CVI and RMI are the processes that display, from our viewpoint, the best potential in terms of cost and volume production. Further, they are complementary, i.e., the residual porosity of CVI-composites, at a suitable state of densification, can be filled via an RMI-step. Conversely, the *PIP-process*, which is also a low-temperature technique, is lengthy since several time-consuming PI/P sequences (from 6 to 10) are necessary to achieve an acceptable densification. It yields composites with a significant residual porosity and implies considerable handling. It can also be combined with RMI, as previously mentioned. Finally, *SI–HP* is both a HT (1700–1800°C for SiC) and a high-pressure (≈ 25 MPa) process, which is only compatible with fibers of high thermal stability (carbon or stoichiometric SiC fibers) with a risk of fiber degradation.[10] It has been improved through the use of nanometric SiC particles slurry and additives (Al_2O_3, Y_2O_3) forming a liquid phase at sintering temperature (see, e.g., the NITE-process).[11] One of the main advantages of SI–HP lies in the fact that it is a fast densification process, yielding composites with almost no residual porosity and hence a high thermal conductivity. However, its extension to large multidirectional fiber preforms seems to be problematic.

From this brief analysis, it appears that none of the existing processes is perfect and that *hybrid* techniques combining two approaches, such as PIP/RMI or CVI/RMI, might presently be the most appropriate choice; each step could still be improved in order to gain in reproducibility and cost, at plant level.

Material Design

The choice of a suitable *reinforcement*, for a given matrix, is dictated by several considerations including FM compatibility, mechanical or/and thermal properties, chemical compatibility with the high service temperature, density, and cost. Covalent nonoxide fibers (carbon and oxygen-free SiC fibers) display the best HT mechanical properties and can be good heat conductors

(depending on their microstructure). Further, they are light and some of them are available in large quantity at a relatively low cost (carbon fibers). Obviously, they are the reinforcement of choice for nonoxide matrices, e.g., SiC, since SiC is in thermodynamic equilibrium with carbon at high temperature. Also, there exist well-identified nonoxide interphases (pyrocarbon and boron nitride) compatible with both carbon and SiC within a wide temperature range. Unfortunately, nonoxide CMCs are oxidation prone and their long exposure to oxidizing atmospheres requires efficient protection against oxidation (PAO). Oxide-based CMCs are by essence inert in oxidizing atmospheres and might appear more attractive. However, most *oxide-based* fibers (containing α-alumina, mullite, or zirconia) display poor HT mechanical properties (they suffer from grain growth and creep beyond about 1000–1100°C). Further, there is presently no stable oxide interphase formally equivalent to pyrocarbon or BN, i.e., a layered oxide with a low shear strength that could be easily deposited on fibers (although there are few oxide interphase materials, such as monazite or hibonite, that could deflect matrix cracks in oxide–oxide composites but not as easily as their nonoxide counterparts).[12] Also, oxide-based CMCs are insulating materials and their density can be slightly higher than that of C- or SiC-based composites. Finally, fibers should exhibit a good weavability, which supposes a low enough diameter (typically, 10 μm or less) when their stiffness is high (this is the case for most carbon fibers but not for all stoichiometric SiC fibers) and preferably a high failure strain. To conclude, nonoxide CMCs (C/SiC or SiC/SiC) are presently preferred for most structural applications even though their use in oxidizing atmospheres raises a difficult problem of durability.

The choice of a concept of *damage tolerance* is a key step in the design of CMCs. In SiC-matrix composites, damage tolerance is achieved through a weakening of the FM-bonding (controlled by an interphase), which allows the matrix microcracks to be deflected by the FM-interfaces. However, another concept of damage tolerance, based on the use of a highly porous matrix (and no interphase), is known. Its use might be appropriate in oxide/oxide composites since both constituents are inert in oxidizing atmospheres.[13,14] Conversely, it might be problematic in nonoxide CMCs since a porous matrix will favor fiber oxidation and lower thermal conductivity.

The design of the interphase in SiC-matrix composites is not straightforward since the interphase has com-plex and contradictory functions.[4] First, it should arrest and deflect the matrix microcracks, which supposes that the interphase debonding energy, Γ_i, is low relative to the failure energy of the fiber, Γ_f, a generally accepted criterion being $\Gamma_i/\Gamma_f < 1/4$.[5] This is the so-called mechanical fuse function (the interphase protecting the fiber from an early failure). Second, the interphase may act as a diffusion barrier (as previously mentioned for the RMI process) and relax partly thermal residual stresses. It has been recently postulated that the best interphase materials might be those with a layered crystal structure or micro-structure, the layers being deposited, parallel to the fiber surface, weakly bonded to one another (for low Γ_i) but strongly adherent to the fiber surface (to avoid debonding at the fiber surface).[4] In SiC-matrix composites, the best interphase material from a mechanical standpoint is probably an anisotropic *pyrocarbon* (Fig. 1a).[4,15] Unfortunately, pyrocarbon is intrisically oxidation-prone at temperatures as low as 500°C. *BN* is an interesting alternative since it has a similar layered crystal structure and a better oxidation resistance, its oxidation starting at about 800°C and yielding a fluid B_2O_3 oxide known for its healing properties. However, its formation on a SiC fiber is not straightforward. When deposited at low temperature by CVD/CVI, it is amorphous or poorly crystallized and hence sensitive to moisture. Its crystallization by heat treatment is often limited by the thermal stability of the fibers and the bonding with the fibers is poor.[16] An interesting alternative might be to form a more adherent BN coating by annealing a SiC fiber containing some boron (used as a sintering aid) in a nitriding atmosphere at high temperature. SiC/SiC composites with such an *in situ* formed BN interphase have been reported to be more oxidation resistant than those with a BN interphase deposited by CVD/CVI.[17]

Since the number of thermally stable materials with layered structures is limited, the concept of layered inter-phase has been further extended to materials with a lay-ered microstructure at the nanometer scale, i.e., to $(X–Y)_n$ *multilayers*.[4] Such interphases offer a much higher design flexibility, the adjustable parameters by, e.g., pressure-pulsed CVI (or P-CVI) being the nature of X and Y, the overall thickness of the interphase, the thicknesses of the X and Y sublayers, the number of $X–Y$ sequences, n, and the X/Y bonding. As an example, in $(PyC–SiC)_n$ or $(BN–SiC)_n$, the amount of oxidation-prone mechanical fuse ($X =$ PyC or BN) can be strongly reduced (the thickness of X-layers being a few nanometers, typically 3–20 nm) with the result that the durability of the composites in
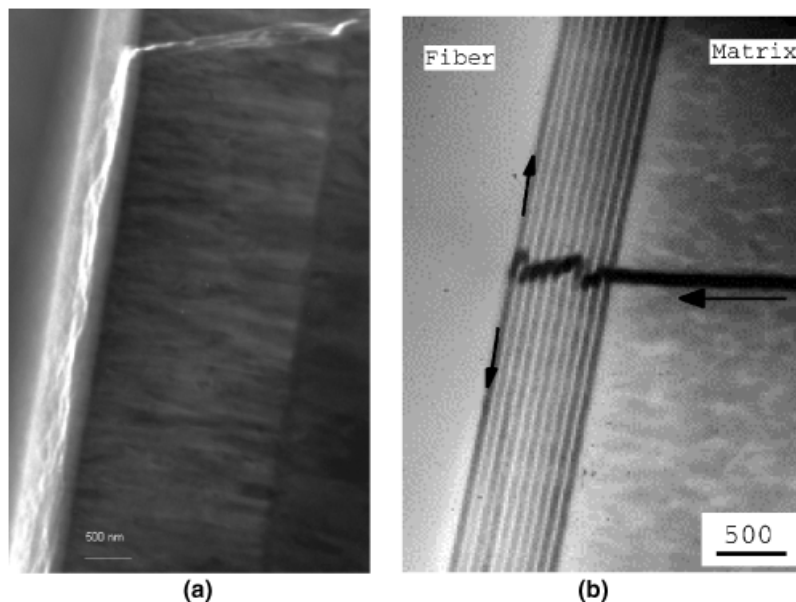
Fig. 1. Interphases for SiC/SiC composites with layered crystal structure or microstructure: (a) anisotropic pyrocarbon single-layer interphase[15] and (b) (PyC–SiC)$_{10}$ multilayered interphase.[18]

oxidizing atmospheres is improved by self-healing phenomena (silica or SiO_2–B_2O_3 scales formed by oxidation healing the narrow annular pore created around each fiber by oxidation) (Fig. 1b).[4,18] Another interphase concept that has been less explored is the use of a *porous* SiC layer, a porous solid displaying a lower failure energy than its dense counterpart. However, such a porous interface would favor the oxidation of the fibers as mentioned previously for porous matrices.

Finally, a *seal-coating* is usually deposited on the external surface of C/SiC and SiC/SiC composites, mainly to seal the residual open porosity (composites fabricated by the PIP or CVI processes) or/and to improve their resistance to corrosive environments. Dense single layer ceramic coatings (such as SiC or Si_3N_4) displaying a tendency to microcracking (as a result of CTE-mismatch or mechanical loading) multilayered coatings are preferable, as it will be discussed in the next section.[19] Such coatings are deposited by PVD or P-CVD.

Selected Properties

Mechanical Behavior

SiC-matrix composites display a nonlinear stress–strain behavior when tensile loaded in one of the fiber directions. This nonlinearity is related to the occurrence of damaging phenomena, mainly including multiple matrix microcracking and FM-debonding. As a result, their stiffness progressively decreases as the applied load is raised beyond the proportional limit (SiC/SiC composites), with little permanent deformation upon unloading (at least for well-processed materials). Hence, they are often referred to as *damageable elastic* materials. The extent of the nonlinear domain in which the materials are damage-tolerant is related to the ultimate failure strain of the fibers (e.g., the latter becoming low, typically 0.6–0.7% for stoichiometric SiC fibers). Further, the damage features are strongly related to the intensity of the FM-bonding, a point that is often underestimated.[15] When the FM-bonding is too weak, the matrix microcrack density is low, the microcracks are widely open under load, and debonding occurs over a long distance (and sometimes over the whole fiber length, exposing the oxidation-prone fibers to the ambient environment). By contrast, when the FM-bonding is stronger and the interphase is strongly adherent to the fiber, it is the reverse situation that is observed, the composite displaying a higher failure stress (Fig. 2) and a better oxidation resistance. SiC-matrix composites are tough when properly designed and fabricated with toughness, expressed in terms of critical energy release rate of the order of 10 kJ/m^2, whereas that of monolithic SiC-ceramics is of the order of a few 100 J/m^2.[15]
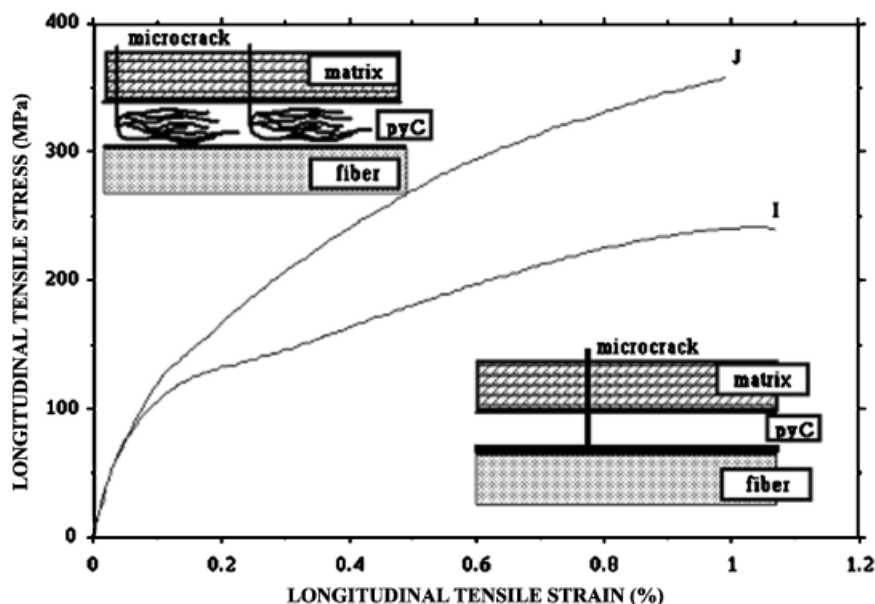
Fig. 2.   Typical stress–strain tensile curves of 2D-SiC(Nicalon)/PyC/SiC composites with weak FM-bonding (material I) and stronger FM-bonding (material J), corresponding to different matrix crack deflection schemes (according to Droillard[15]).

Finally, they are less fatigue-prone than metals and alloys with stress threshold below which no fatigue failure occurs, of the order of 75% the ultimate failure stress under static loading.[20]

The tensile stress–strain behavior of SiC-matrix composites does not change markedly up to $\approx 1100°C$. However, some change may be observed either at higher temperatures if the fibers are limited in thermal stability (case of the unstable Si–C–O fibers) or even at lower temperatures when an oxidizing atmosphere has access to the fibers and the interphase (case of insufficiently protected materials). Further, SiC-matrix composites creep at high temperatures with a creep rate depending on the nature of the fibers (stoichiometric microcrystalline SiC fibers prepared or treated at high temperatures being more creep-resistant than their Si–C–O nanocrystalline counterparts) and that of the matrix.[21]

Thermal Conductivity

Thermal conductivity is a key property in many HT applications of CMCs. Generally speaking, SiC-matrix composites are relatively good conductors of heat but their thermal conductivity depends on the crystallinity of their constituents, the FM-bonding, and residual porosity. The thermal conductivity of carbon fibers can be very high $(100 \, W/m \cdot K$ and more) for those fibers fabricated from mesophase pitch and heat treated beyond $2500°C$ (P55–P130 series) but low $(10 \, W/m \cdot K$ and less) for poorly organized fibers (ex-PAN T300 fibers).[22] In a similar manner, nearly stoichiometric SiC fibers fabricated at high temperatures (e.g., Tyranno SA fibers, Ube Industrial, Japan) display a much better conductivity than the quasi-amorphous Si–C–O fibers prepared at low temperatures, typically 65 and $10 \, W/ m \cdot K$ at room temperature, respectively.[23] Equally important is the effect of the residual porosity, a composite produced by RMI or hot pressing $(V_p \leq 5\%)$ exhibiting a higher conductivity than a composite fabricated by PIP or CVI $(V_p \approx 10–15\%)$. Hence, a SiC/SiC composite is expected to show a thermal conductivity of the order of $30 \, W/m \cdot K$ at $1000°C$ when prepared from nearly stoichiometric SiC fibers with almost no residual porosity, and possibly higher if the reinforcement consists of graphitized carbon fibers (with, however, in this case a risk related to the occurrence of microcracking due to CTE-mismatch that will lower the conductivity).

Oxidation Resistance

In most thermostructural applications, SiC-matrix composites are exposed to oxidizing atmospheres. Since their constituents are intrinsically oxidation-prone, their behavior under such environments is of key importance
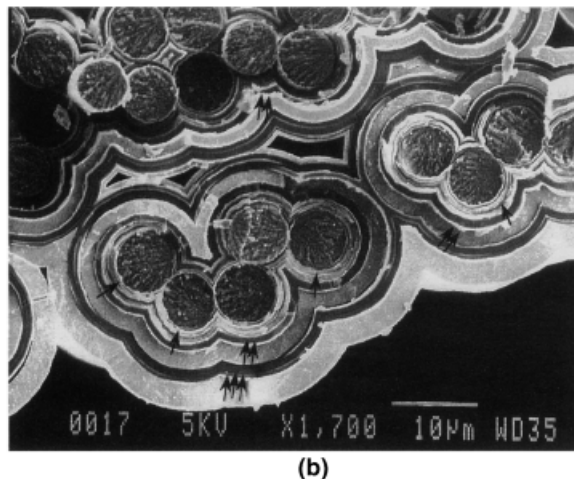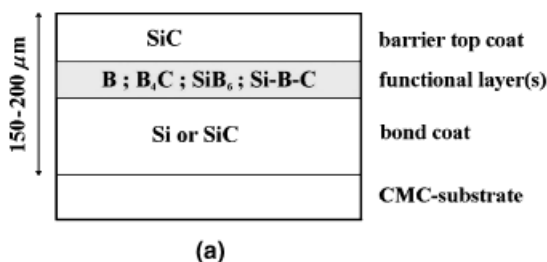
Fig. 3.   Nonoxide CMCs with improved oxidation resistance through the use of multilayered seal-coating (a) and multilayered self-healing matrix (b). Adapted from Goujard and Vandenbulcke[19] and Lamouroux et al.,[27] respectively.

for durability. When a SiC/SiC (or a C/SiC) composite with a pyrocarbon (or BN) interphase is heated in an oxidizing atmosphere, *active or passive* oxidation phenomena are observed depending on whether all the reaction products are gaseous (CO or CO_2 for carbon, CO and SiO for SiC) or at least one reaction product is condensed as a covering protective scale (silica for SiC and boria for BN), respectively. Fortunately, in many cases the oxidation regime is passive. Under this assumption, the effect of oxidation on the microstructure, mechanical, and thermal properties depends on the oxidation conditions (temperature, oxygen partial pressure) and material parameters (interphase thickness).[24] At *low temperatures*, $500 < T < 900°C$, the kinetics of oxidation of the pyrocarbon interphase in a SiC/PyC/SiC composite is already fast whereas that of SiC is almost negligible. As a result, oxidation is an in-depth phenomenon that progressively consumes the interphase, destroys the FM-bonding, degrades the mechanical behavior, and alters the thermal conductivity. The effect is still more significant if the composite is reinforced with carbon fibers and heavily microcracked as a result of an applied load or CTE-mismatch.[25] Conversely, at high temperatures (1000–1200°C) the kinetics of formation of silica (and boria when a BN-interphase is used) is fast and the condensed oxide scale (which is covering for both B_2O_3 on BN and silica on SiC) is protective and tends to seal or/and fill the residual pores and microcracks, stopping (or at least slowing down) the in-depth diffusion of oxygen. Hence,

the behavior of C/SiC and SiC/SiC composites in oxidizing atmospheres is usually better at relatively high temperatures ($\approx 1200°C$) than at lower temperatures. However, this protection due to condensed oxides is insufficient for long exposures under load and even disappears in wet atmospheres (volatilization of silica).[26] Under such conditions, some protection against oxidation is necessary.

There are two ways to improve the *oxidation resistance* of SiC-matrix composites, which are based on multilayered seal-coatings or self-healing matrices. Homogeneous single-layer coatings, such as dense SiC-coating, provide an insufficient oxidation protection for C/SiC and SiC/SiC composites submitted to thermal shocks or/and mechanical cyclic loading. In both cases, microcracks are formed in the coating that favor the in-depth diffusion of oxygen. A first strategy is to use a *multilayered seal-coating* that usually consists of the following: (i) a bond coat, such as a dense layer of SiC (for SiC CVI-matrix) or silicon (for SiC+Si RMI-matrix) deposited on the external surface of the composite at the end of the fiber preform densification, (ii) a functional layer containing species (such as B, B_4C, SiB_6, or Si–B–C mixture) that can form fluid oxides (B_2O_3 or B_2O_3–SiO_2) when exposed to an oxidizing atmosphere in a given range of temperature, and (iii) a barrier top-coat that can be a dense SiC-layer, the overall thickness of the coating being of the order of 150–200 μm (Fig. 3a). When the coating undergoes microcracking upon cyclic loading, the microcracks are being filled by the

fluid oxides and the in-depth diffusion of oxygen is slowed down or stopped.[19] An even more efficient strategy consists in replacing the homogeneous SiC-matrix itself by an engineered multilayered matrix based on a similar principle, referred to as a *self-healing matrix* and that could be infiltrated by P-CVI at the laboratory scale.[27] Here, the matrix is deposited as the repetition of a given sequence S comprising thin layers *X* acting as mechanical fuses (*X* being C, C (B), BN, BN (Si), or any suitable fuse) and layers *Y* of species forming fluid oxides, as mentioned above (Fig. 3b). Durability of the order of 1000 h under cyclic loading in air up to 1100°C has been reported for SiC fiber composites fabricated with such self-healing matrices.[28] Finally, specific multilayered coatings containing oxide layers (such as mullite or/and baryum strontium aluminosilicate, BSAS) have been proposed and tested for SiC/SiC composites exposed to *wet* oxidizing atmospheres to reduce the recession rate of the materials, e.g., in hot combustion gas rich in water vapor.[29]

Representative Applications

Space and Aeronautic Field

SiC-matrix composites are potential material candidates for the fabrication of *hot structures* of spacecraft, as demonstrated at the prototype part level some years ago within the scope of the Hermes European space shuttle project.[30] Here, the maximum temperature ranges from 800°C to 1600°C, during the ascent and re-entry phases of a flight, the structures being submitted to thermal shocks and cyclic mechanical loading under ablative or passive oxidizing atmospheres, with an expected durability of a few tens of hours (Fig. 4a). Such conditions are compatible with modern composites fabricated with carbon fibers (to reduce weight and achieve good mechanical properties at the highest temperatures) and engineered multilayered self-healing matrix. However, some environmental barrier coating (EBC) might be necessary to limit the recession rate of the material under active oxidation/ablation regime (HT and velocity combined with low $P(O_2)$). Such a material approach will benefit from the high refractoriness of C/SiC composites (≈ 2500°C) relative to the low melting point of aluminum (≈ 650°C) in the metallic option combined with a thermal insulation. Further, the CVI-process is well suited to the fabrication of large size structures (two meters or more), its feasibility already

established and joining techniques under development. Similar materials could also be used for the shielding of satellites against the impact of meteorites or foreign objects owing to the high toughness and hardness of these materials.

Another promising field of application is that of the hot structures of *aerojet engines* and related *gas turbines*, which are presently made of heavy, low melting point nickel-based superalloys that require complex cooling systems. Replacing superalloys by SiC-based composites would permit to raise the gas temperature, suppress, or at least limit the cooling requirement, increase the efficiency of the engine, and reduce both the weight and the noise/pollution level (Fig. 4b). However, it will probably take some time (these materials are still young and their fabrication costly) and be limited in a first step to nonrotating parts, i.e., the combustors and the afterburner parts such as the flaps of the exhaust nozzles. The main concern here is *durability*, that should be of the order of several thousands of hours. The outer (divergent) flaps of exhaust nozzles experience a temperature that is relatively low ($T < 700$°C). Hence, they can be fabricated with carbon fiber-reinforced SiC-matrix, with a weight gain of 50%. Tested in flight as early as 1989, they are now in volume production (M88 Snecma engines of the Rafale fighter, Paris, France).[28,30,31] The inner (convergent) flaps of the exhaust nozzles are exposed to higher temperatures (up to 1100°C). Durability of the order of 1000 h has been demonstrated for 3D-composites with a self-healing multilayered matrix on the basis of bench combustion tests. The next step is the combustion chamber or combustor, whose fabrication with SiC-based composites is in progress. No results of tests are presently available for military aerojet engine combustors in the open literature, as far as we know. However, the use of SiC-matrix combustor in power plant gas turbine of cogeneration is well documented, with similar (not to say more severe) service condition.[32,33] Combustors of large size comprising concentric cylindrical CMC liners have been fabricated by CVI or RMI with SiC (Hi-Nicalon)/BN/SiC (Si) composites. Durability of several 10,000 h has been established under real service conditions, for composites with a BSAS–EBC.[29] To conclude, the use of SiC-matrix composites in the hot nonrotating parts of gas turbines (aerojet engines and cogeneration gas turbines) appears to be promising. It is now a matter of engineering, reliability, and cost (that of performant SiC fibers still remaining relatively dissuasive).
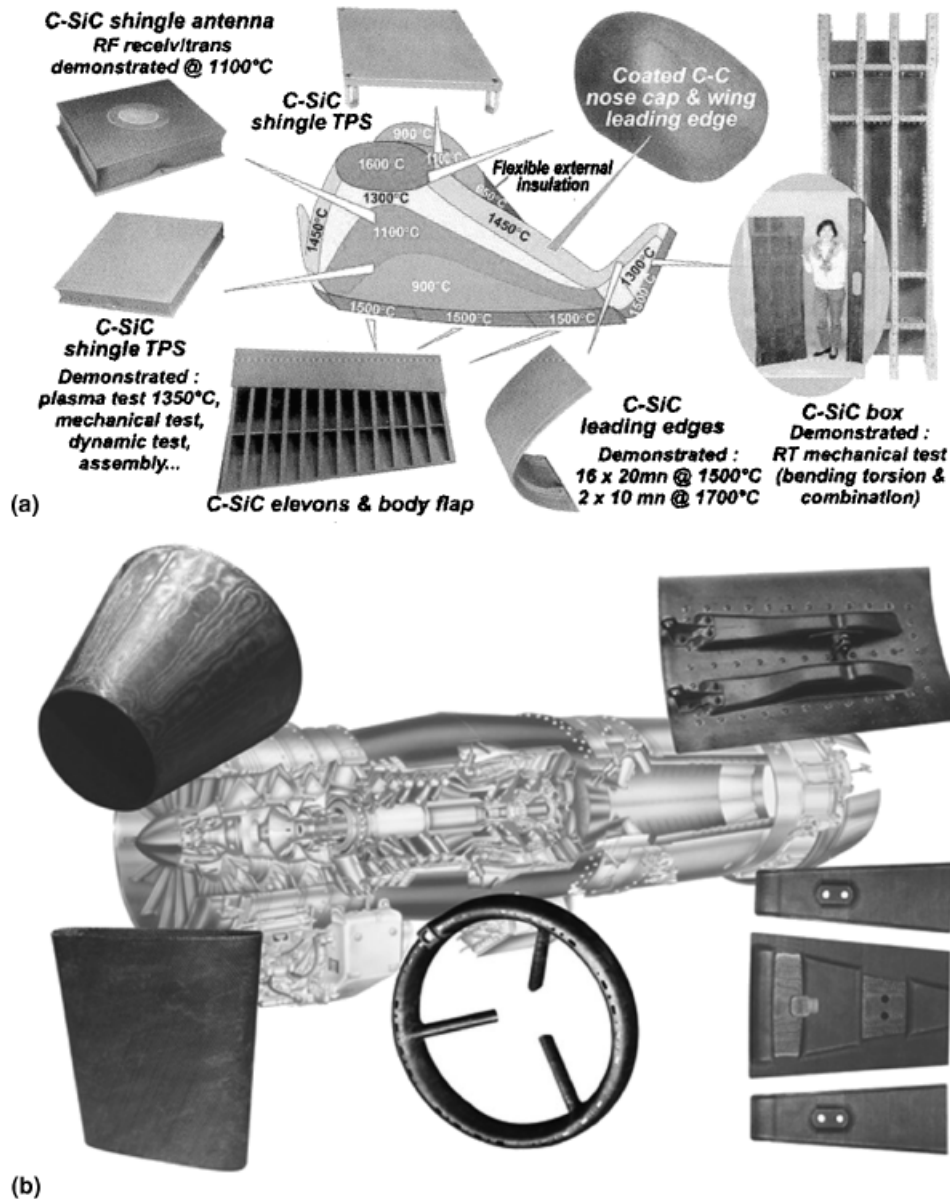
Fig. 4.   Potential applications of SiC-matrix composite materials: (a) in spacecraft hot structure and (b) in aerojet engine. Adapted from Christin.[30]

Braking systems is another field of applications. Up until recently, aircraft and racing car disk brakes were fabricated with C/C composites sliding against themselves (disk/disk or disk/pad configuration). C/C brakes display, relative to conventional steel disk brakes, a lower density, a higher service temperature (and hence improved security), good friction properties at HT, and longer lifetime. Conversely, their friction properties

near the ambient temperature are altered in wet atmosphere and their wear is significant. It has been proposed recently to replace part of the C-matrix by a SiC-based matrix, the new matrix being formed by a combination of carbon-PIP and SiC (Si)-RMI techniques, as depicted previously.[34,35] Coated C/SiC (Si) brake disks, sliding against themselves or pads (organic or metallic), show a higher coefficient of friction less depending on moisture

and a very low wear. They are now proposed as an option for cars in Europe and could be extended to other fields (trains, lifts, etc.).

Finally, SiC-matrix composites might become key structural materials in high temperature *nuclear reactors* of the future (e.g., the tokamak fusion reactors in which heat generated by the fusion reaction is extracted through the wall of the toroidal plasma chamber with a cooling agent such as gaseous helium, at a temperature of 800–1000°C), on the basis of their refractoriness, HT-mechanical properties, high thermal conductivity, and more importantly low activation under radiation.[36–39] This domain is becoming an active field of research, the main concerns being: the combined effect of temperature and radiation on the structure (with either a shrinkage or swelling) and mechanical properties of both SiC fiber and matrix, the nature of the interphase (thick layers of carbon or BN being inappropriate under radiation), the effect of activable impurities in the composites, the thermal conductivity, and the corrosion by residual or *in situ* formed gaseous species (oxygen or helium in fusion reactors). It appears from preliminary data that the composites should be better fabricated with fibers and matrix consisting of crystalline β-SiC with a low impurity content (additives introduced as sintering aids in fibers or/and matrix being a subject of concern), with either porous SiC, $(PyC–SiC)_n$, or $(BN–SiC)_n$ multilayered interphases of low C or BN content, and displaying a low residual porosity (for hermeticity and thermal conductivity considerations). Further and as mentioned previously, large size structures could be fabricated by already existing technologies (PIP/RMI or CVI/RMI combined techniques and joining).

Conclusion

SiC-matrix composites, i.e., C/SiC and SiC/SiC, can be fabricated from different carbon or SiC fibers displaying a variety of properties, usually by single (such as CVI) or combined processes (e.g., PIP/RMI or CVI/RMI). They are damage-tolerant when the fibers and matrix are bonded together with an interphase that can be a single relatively thick layer of pyrocarbon (or BN), a porous single layer of SiC, or an engineered multi-layered $(X–Y)_n$ interphase (with $X = PyC$ or BN and $Y = SiC$) offering more design flexibility. The composites display a nonlinear stress–strain behavior under tensile loading related to damaging phenomena at the

microstructure scale, a toughness of the order of 10 kJ/m^2, and a good fatigue resistance. They creep at high temperature but with a creep rate that is lower than that of metals. Their thermal conductivity is relatively high when their constituents are well crystallized and their residual porosity is low. Their oxidation resistance is better at 1000–1200°C than at lower temperatures due to the formation of protective oxide scales (SiO_2, B_2O_3), in the so-called passive regime. It can be improved with multilayered self-healing matrices and coatings, based on the formation of fluid healing oxides. The oxidation resistance is degraded in wet atmospheres where specific EBCs needed to be used.

SiC-matrix composites are matured enough to be utilized in a variety of applications including the hot structures of spacecraft, aerojet engines, and gas turbines of cogeneration, some parts being already in volume production, with a weight gain of $\approx 50\%$ versus superalloys and a durability that can be several thousands of hours. They are also promising materials for braking systems and HT nuclear reactors of the future.

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