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Research progress on silicon carbide and its modified coatings in C/SiC composites

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Abstract: Carbon fiber reinforced silicon carbide ceramic composite (C/SiC) has become a key structural material due to its excellent high temperature resistance, corrosion resistance and oxidation resistance. However, C/SiC composites are prone to oxidation under long-term high temperature loading conditions. In this work, the research progress of SiC coating and its modified coating on the surface of C/SiC composite is reviewed. The optimization of these coatings mainly involves two aspects: structure and composition. The focus of structural optimization is to improve the interphase structure by such as SiC nanowires, thereby improving the interfacial bonding strength between the coatings and between the coatings and the substrate. The focus of composition optimization is to improve the performance of the coatings under high temperature loads by rare earth silicates, etc. The modification strategies of various modified coatings are emphatically introduced, which is helpful to guide the preparation of high-performance C/SiC coating materials in the future.

Key words: C/SiC composite; coating; coating modification; oxidation; strength

1 Introduction

Carbon fiber reinforced silicon carbon ceramic composite (C/SiC) is a structural material with excellent high temperature resistance, corrosion resistance and excellent mechanical properties. Due to its high strength and hardness, C/SiC composite can effectively reduce its mass while extending the service life of the structure. The toughness of the carbon fiber phase and the strength of the silicon carbide phase make the whole material have better thermal shock damage resistance and mechanical stability under complex working conditions. C/SiC composite is mainly used in vehicle braking systems and photovoltaic thermal field components [1−3].

As a structural material under complex working conditions, the damage of C/SiC composites is mainly due to mechanical erosion damage and oxidative chemical damage [4−6]. Despite the inherently excellent properties of C/SiC composites, thermal gradients experienced during thermal shock can result in matrix cracking, complete debonding between the matrix and fibers, as well as various forms of multiple cracking phenomena. The damage mechanisms associated with thermal shock encompass fiber oxidation failure, matrix damage, and debonding at the C/SiC composites interphase. During instances of thermal shock, stress−strain cycles reveal an evolution in mechanical properties corresponding to the number of cycles applied; notably, the strength of C/SiC composites diminishes with increasing temperature

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and a greater number of thermal shock cycles. Furthermore, oxidation within C/SiC composites is concentrated along the boundary between active and passive oxidation [7−11]. As cracks develop within the material, a porous and discontinuous glassy silicon oxide film emerges on the surface of C/SiC composites throughout this damaging process [10,12,13].

During application, as the temperature and mechanical loading increase, the composite material tends to loosen, leading to the formation of cracks in the SiC matrix on its surface. These cracks can directly facilitate the oxidation of carbon fibers and may even result in debonding at the interphase between fiber bundles and the matrix. This phenomenon significantly diminishes mechanical properties and ultimately leads to material failure due to oxidation. The preparation of coatings on the surface of C/SiC composites has emerged as a crucial strategy for enhancing their oxidation resistance. The coating serves two primary functions: firstly, it helps to maintain certain mechanical properties of C/SiC composites during oxidative corrosion; secondly, it can be through the coating in the oxidation process reaction products wrapped in the crack defects of the material, or through the coating to deflect the cracks generated in the matrix, thereby preventing the formation of oxygen diffusion channels [14−19].

The development of protective coatings for C/SiC composites has resulted in the emergence of various types of coatings. These coatings can primarily be categorized into two major groups: carbide series coatings and silicate coatings. The carbide series coatings encompass SiC coatings, ZrC coatings, and other related carbide coatings [20−22]. Silicate coatings include aluminum silicate coatings as well as rare earth silicate coatings [23,24]. Furthermore, there exist additional types of coatings, such as silicide and metal-based coatings [25,26].

2 SiC coatings

C/SiC composites are susceptible to oxidation at elevated temperatures, which adversely affects their mechanical properties. The application of SiC coatings not only safeguards the C/SiC composite matrix from oxidative degradation but also enhances the material's strength. For instance, these coatings are utilized in the aerospace sector to protect engine components, in the automotive industry to shield brake disc components, and in the energy sector to safeguard gas turbine and nuclear reactor components. SiC coatings serve as an effective protective measure for C/SiC composites across a diverse array of applications.

SiC exhibits exceptional properties, including a good corrosion resistance, a low thermal expansion coefficient, a high melting point, and a remarkable strength. Furthermore, its thermal expansion coefficient and physical-chemical characteristics closely resemble those of carbon−ceramic matrix materials. Consequently, SiC is the primary substance utilized as a C/SiC composite coating. The application of a SiC coating significantly enhances both the mechanical properties and oxidation resistance of C/SiC materials.

2.1 Development history

SiC coatings have advanced alongside the development of carbon materials. Initially, SiC coatings were applied during the fabrication of carbon−carbon preforms for C/SiC composites and were first investigated for their oxidation resistance in carbon fiber reinforced carbon (C/C) composites. With ongoing advancements in material science, the preparation of C/SiC composites was successfully achieved in the 1970s. This period also marked the beginning of research aimed at enhancing the fiber−matrix interphase through the integration of carbon fibers with SiC coatings.

Since the 1990s to the present day, with the widespread application of carbon−ceramic composite materials, a key focus has been on preparing dense and uniform coatings that can be controlled for their density and uniformity. In the future, according to different application scenarios, new requirements for different working conditions will be met by preparing different SiC modified coatings.

2.2 Preparation methods

SiC and its associated modified coatings are commonly prepared utilizing various methods such as chemical vapor deposition (CVD), atmospheric plasma spraying (APS), sol−gel techniques, and slurry coating-sintering processes. Spraying methods, specifically APS, offer precise control over both the coating thickness and the bonding strength between the coating and the substrate. However, they necessitate specialized equipment for effective operation. In contrast, sol−gel method entails the preparation of a precursor followed by the formation of the coating through high-temperature heat treatment. This technique is characterized by low costs and results in coatings that exhibit good uniformity. Nevertheless, these coatings are susceptible to cracking and tend to be thinner than those produced by other means. Slurry coating-sintering approaches provide convenience while allowing for flexible adjustments of the slurry composition. However, controlling coating thickness and ensuring uniformity remain challenging aspects of this methodology [27].

CVD is the most widely employed technique for the preparation of SiC coatings, which is highly controllable, and resulting in coatings that exhibit excellent purity and adhesion properties, thereby providing effective protection for C/SiC composites. The deposition rate during chemical vapor deposition significantly influences the characteristics of the coating. By controlling and reducing the deposition rate, it is possible to eliminate gaps within the coating and reduce the formation of defects to some extent. This approach enhances the integrity of C/SiC materials and improves the protective efficiency of the applied coating [19,28,29].

Pulse chemical vapor deposition (PCVD) is capable of producing dense coatings that effectively fill the defects present in the substrate, thereby enhancing the bonding strength between the coating and the substrate. As the number of deposition layers increases, thermal stress within the coating can be alleviated more efficiently, leading to a reduction in both the quantity and width of cracks. Furthermore, an increase in the number of deposition cycles contributes to the improved oxidation resistance in C/SiC composites [30−33].

Oxidation of materials is primarily associated with the development of open pores and cracks within the coating. SiC coatings have been shown to enhance the oxidation resistance of C/SiC composites to a certain extent. However, during their preparation, SiC coatings encounter two categories of defects: support defects and processing defects, specifically matrix voids that occur between fiber bundles after machining. In multilayer SiC coatings, two distinct types of deposition defects can be identified: planar defects located between layers and network defects found at particle boundaries. Increasing the thickness of the coating may effectively reduce processing defects; however, support defects remain unresolved as they are not eliminated by multiple depositions. Increasing the deposition temperature and decreasing the deposition pressure helped to remove planar defects. It has been observed that when multilayer films are free from planar defects, preparation-related imperfections can be entirely sealed off [34,35].

2.3 Shortcomings and ways

Although the SiC coating significantly enhances the performance of composite materials, there exist several shortcomings. The primary concerns include the gradual degradation in prolonged high-temperature oxidation environments, insufficient bonding strength between the coating and the substrate that can lead to interfacial delamination and crack propagation, as well as a pressing need for further improvements in the stability and durability of the coating under thermal shock and complex corrosion conditions. The bonding strength between the coating and the fiber or substrate exhibits significant variation depending on the method employed for preparing the SiC coating. Consequently, it is essential to select and refine the coating preparation process in accordance with different coating compositions. For instance, pressure adjustment can be implemented during chemical vapor deposition to enhance both the density and uniformity of the coating. However, if the coating continues to exhibit inadequate resistance to external environmental corrosion under prolonged oxidation conditions, it becomes imperative to optimize its composition and microstructure, as well as design an interphase layer aimed at improving material performance [36−38].

Specific methods include: First, the overall design of gradient and multilayer composite coatings is employed to control microstructure, mitigate interfacial defects, and enhance interlayer bonding. Second, the incorporation of modified materials aims to improve specific properties of the SiC coating. For instance, doping components can form a three-dimensional protective film with SiC oxidation products at specific temperatures. Under thermal shock conditions, these modified components can bolster mechanical strength and enhance the overall thermal shock stability of the material. Furthermore, corrosion-resistant modified components can be added to enhance the material's corrosion resistance in complex environments.

2.4 SiC interphase coatings

2.4.1 SiC deposited carbon fibre interphase

Carbon fibers are commonly used as toughening phases in C/SiC composites, and they are often coated using CVD for pyrolytic carbon (PyC) deposition. The pyrolytic carbon serves as an intermediate buffer layer located between the carbon fibers and the SiC matrix [39]. The SEM images of pyrolytic carbon coated carbon fibers are shown in Figs. 1(a, b) [40]. The pyrolytic carbon exhibits good adhesion to the carbon fibers and prevents direct contact between the carbon fibers and the matrix. When the composite material is damaged, the presence of the pyrolytic carbon buffer layer leads to fiber pull-out, thereby limiting damage propagation and ultimately enhancing the overall performance of the composite material [40−42].

Similarly, SiC interphases can be deposited onto carbon fibers to replace the pyrolytic carbon layer. By depositing a layer of SiC interphase on the fibers through CVD, the interphase is continuous and uniform, with good surface adhesion, filling the surface defects of the fibers and making the fibers overall smooth. Notably, after a deposition period of 1−2 h, the mechanical properties of C/SiC composites are significantly enhanced, with optimal encapsulation effects observed [43,44]. The cross-section micrographs and surface micrographs of CVD-SiC coated PIP-C/SiC composites at different deposition time are shown in Figs. 1(c−h) [44].

After high-temperature annealing, carbon fibers with SiC interphases continue to exhibit significant tough fracture behavior at the interface between the two materials, which is advantageous for enhancing the mechanical properties of the

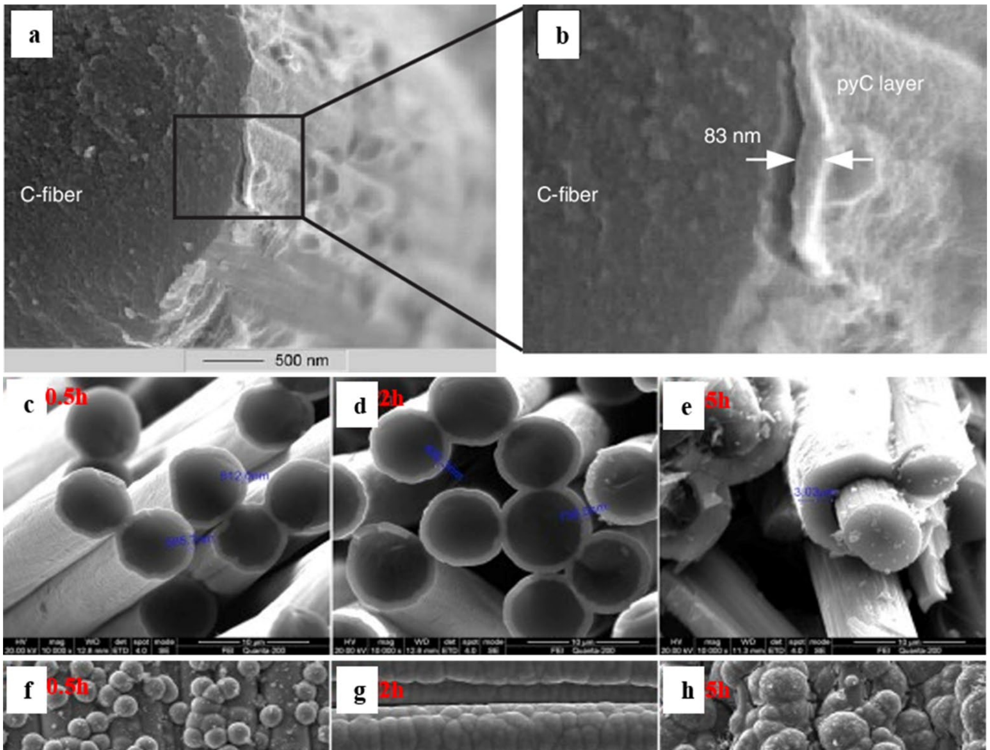


Fig. 1 SEM images of pyrolytic carbon coated carbon fibers (a, b) (Reproduced with permission from Ref. [40]), and cross-section (c−e) and surface (f−h) micrographs of CVD-SiC coated PIP-C/SiC composites at different deposition time (Reproduced with permission from Ref. [44])

composite. However, during the thermal shock processes, partial damage occurs at the interface between the carbon fibers and SiC due to their different thermal expansion coefficients. This discrepancy leads to a reduction in mechanical properties and compromises the integrity of the composite material. High-temperature continuous oxidation can be investigated through thermogravimetric analysis to elucidate the oxidation mechanism of SiC interphases on carbon fibers. At specific temperatures (e.g., 1400 °C), both the SiC interphase and continuous adherent vitreous silica layer formed provide protective effects for the carbon fibers; this represents a passive self-repairing form of oxidation for the interphase. However, when temperatures exceed 1500 °C, substantial differences in thermal expansion coefficients of vitreous silica between high temperature and room temperature result in crack formation. These cracks indicate oxygen channel development, thereby accelerating failure at the interphase until the exposure of carbon fibers to environmental conditions occurs [45−47].

2.4.2 SiC deposited C/SiC coating

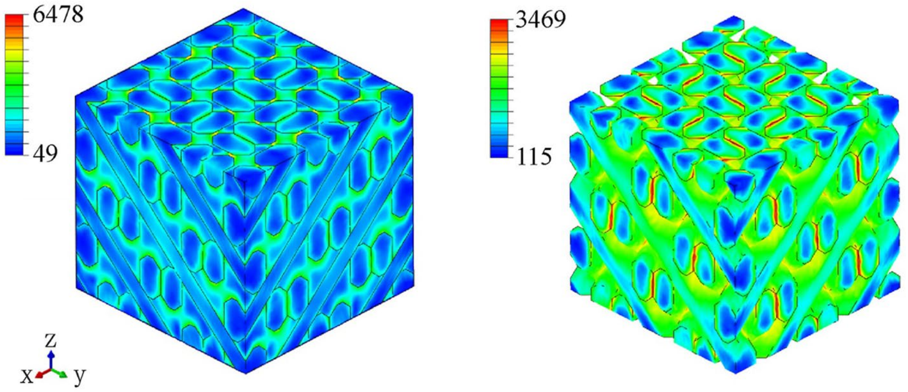
Through a comparative analysis of various coatings, it is evident that SiC coating emerges as the preferred option for the repair of surface mechanical cracks. This choice not only enhances the crack initiation stress in oxidizing environments but also ensures the preservation of high residual strength and mechanical properties [48−50]. In the absence of the SiC coating, C/SiC composites are susceptible to interphase debonding and fiber oxidation failure due to the thermal expansion mismatch-induced interphase stress [36]. Conversely, when a SiC coating is present, the interphase stress and carbon monoxide gas generated at the interface work synergistically to form a vitreous silica oxide film. Initially, this oxide film exhibits a degree of fluidity that allows it to seal surface defects and cracks, thereby creating an effective barrier that delays the oxidation of C/SiC composites. However, during the oxidation process, the silicon in the coating will gradually evaporate, providing a pathway for oxygen diffusion, accelerating the failure of the coating [9,27,51].

The high-temperature damage mechanisms of SiC coatings can be categorized into two primary types: thermal-mechanical erosion and thermochemical corrosion. The first mechanism is thermal-mechanical erosion, which occurs under conditions of severe thermal shock and shear stress induced by high-speed combustion gases, leading to the detachment of portions of the SiC coating and matrix from the substrate. The second mechanism, thermochemical corrosion, during which SiC coating and the matrix passively are oxidized to form a layer of silicon dioxide, which is then corroded into highly volatile $Si(OH)_4$ gas under the action of a large amount of water vapor [4−6].

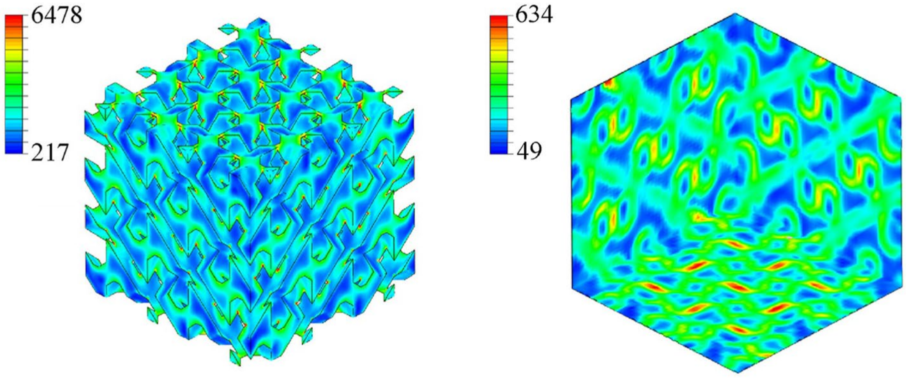
When SiC coating is applied to C/SiC composites, the most critical location during thermal shock occurs in areas of concentrated thermal residual stress, particularly at the interphase between the fiber and the matrix. This vulnerability is more pronounced than that at the interphase between the matrix and the coating. Residual stress levels are influenced by fiber orientation; initial cracks tend to nucleate within the matrix due to higher residual stresses aligned with fiber direction. When deposition direction of the SiC coating is parallel to that of fibers, any delamination that initiates will likely propagate along this same fiber direction [52,53].

The failure and delamination of the SiC coating can lead to severe oxidation failure of the carbon fibers. The primary function of these coatings is to mitigate surface defects on matrices while preventing oxygen diffusion towards fibers. This dual-role not only enhances the protection for matrices but also capitalizes on their superior hardness compared to C/SiC matrices, thereby providing pathways for crack deflection which ultimately improves fracture toughness and strength [48,50,54−57]. The thermal residual stress (TRS) distribution is shown in Fig. 2 [56].

The coating not only exhibits antioxidant properties but also plays a crucial role in preserving the mechanical characteristics of the material, among other functions. Thermal shock does not lead to a significant reduction in the strength of C/SiC composites, and there is no observable change in the toughness fracture mode or bending properties between the fibers and the matrix. The SiC coating effectively regulates the high-temperature flexural strength of C/SiC composites at 1200 °C. The flexural strengths of both uncoated and coated composites decreased from an initial value of 655.3 to 330.7 and 531.2 MPa, respectively [15,18,58].



a. C_f composite b. C_f fiber yarns



c. SiC matrix d. SiC coating

Fig. 2 Thermal residual stress (TRS) distribution of different samples (MPa) (Reproduced with permission from Ref. [56])

3 SiC modified coatings

After the oxidation of C/SiC composites, the SiC coating at the penetrating thickness reveals a carbon fiber combustion zone. This observation indicates that the SiC sealing coating on the surface of C/SiC composites is inadequate for fully protecting the carbon fibers from oxidation. The closure of microcracks results in a decrease in oxygen partial pressure, which exacerbates the oxidation of the SiC matrix adjacent to the carbon fiber combustion zone within C/SiC composites. To mitigate this issue, modifications to the SiC coating are necessary to reduce their degree of oxidation. On the one hand, such modifications can be achieved by employing SiC coatings with varying morphologies; on the other hand, they can also be accomplished through modification using materials with different properties [17].

An effective interface modification is to prepare Si/SiC coatings on the C/SiC substrate, therefore forming high interfacial bonding strength and dense structure. To ensure an adequate supply of silicon, the silicon can react with residual carbon to form an intermediate layer characterized by strong bonding strength. This coating not only exhibits excellent mechanical properties but also demonstrates enhanced wear resistance. However, its oxidation resistance is limited, particularly at temperatures exceeding 1400 °C, where volatilization and re-oxidation of silicon dioxide significantly diminish its oxidation resistance [59−61].

3.1 SiC whisker or nanowire modified coatings

3.1.1 SiC whisker modified coatings

The SiC coating contributes to the surface strengthening by increasing the stress required for crack initiation and enhancing overall strength. The incorporation of SiC whiskers into the SiC coating significantly improves the crack propagation resistance of the matrix, thereby rendering the matrix material tougher and mechanically stronger. This reinforcement effect within the matrix serves as a primary driving force. Both SiC and SiC_w/SiC coatings effectively enhance the density and mechanical properties of C/SiC composites, with the toughening effect of SiC_w/SiC coatings being more pronounced than that observed in pure SiC coatings [50].

Additionally, the incorporation of the SiC whiskers as a strategy to enhance oxidation resistance, in conjunction with corundum aluminum phosphate particles (c-$AlPO_4$) and mullite, results in the formation of a composite environmental thermal barrier coating. The inclusion of SiC whiskers significantly enhances the thermal shock resistance of the coating, mitigating discrepancies in thermal expansion coefficients between the mullite-based coating and adjacent layers, thereby improving the coating's resistance to the crack propagation [62]. Figure 3 [63] shows SEM images and schematic diagrams of SiC whisker to improve the thermal shock properties of materials. The incorporation of c-$AlPO_4$ significantly improves the performance of the coating. During thermal shock, the oxidation of SiC whiskers leads to the formation of a glassy silicon dioxide layer that adheres to both c-$AlPO_4$ and mullite particles. This interaction results in a robust and self-healing aluminum silicate glass structure, which enhances the bonding strength between the coating and the substrate. Consequently, this improvement augments the self-healing capability of the coating, collectively enhancing the material's resistance to thermal shock [64]. Although increasing the thickness of the coating improves the material's oxidation resistance, there exists a threshold beyond which the presence of c-$AlPO_4$ can lead to the generation of PO_x gas during the oxidation process. This phenomenon renders the coating susceptible to defects, such as cracks and voids. Once these defects manifest within the coating, they create penetrating flaws that result in a significant reduction in oxidation resistance [64−69]. The schematic diagram of c-$AlPO_4$ and SiC whisker improving the bonding strength of the c-$AlPO_4$−SiC_w−mullite coating is shown in Fig. 4 [63].

3.1.2 SiC nanowire modified coatings

Similar to the structure and function of SiC whiskers, SiC nanowires represent another significant structural component. The primary method for their synthesis involves the formation of pores through the reaction between SiO_2 and carbon fibers, followed by in-situ growth of SiC nanowires within these pores using CVD. These nanowires are subsequently integrated with SiC or silicon coatings that are also prepared via CVD [70]. The incorporation of SiC nanowires can significantly enhance the bonding strength between the matrix and the coating, improve the tensile strength of the coating itself, and strengthen interfacial bonding performance with C/SiC composite materials. The root-like architecture formed by these nanowires and pores effectively elucidates these enhancements. Notably, the tensile strength of the coating is elevated to over 4.67 MPa, while that of the SiC_{nws}−SiC coating for C/SiC composite materials is from 4.5 to 4.6 MPa.

After undergoing nine thermal cycling processes, this root-structure significantly enhances the oxidation resistance of the coating by alleviating

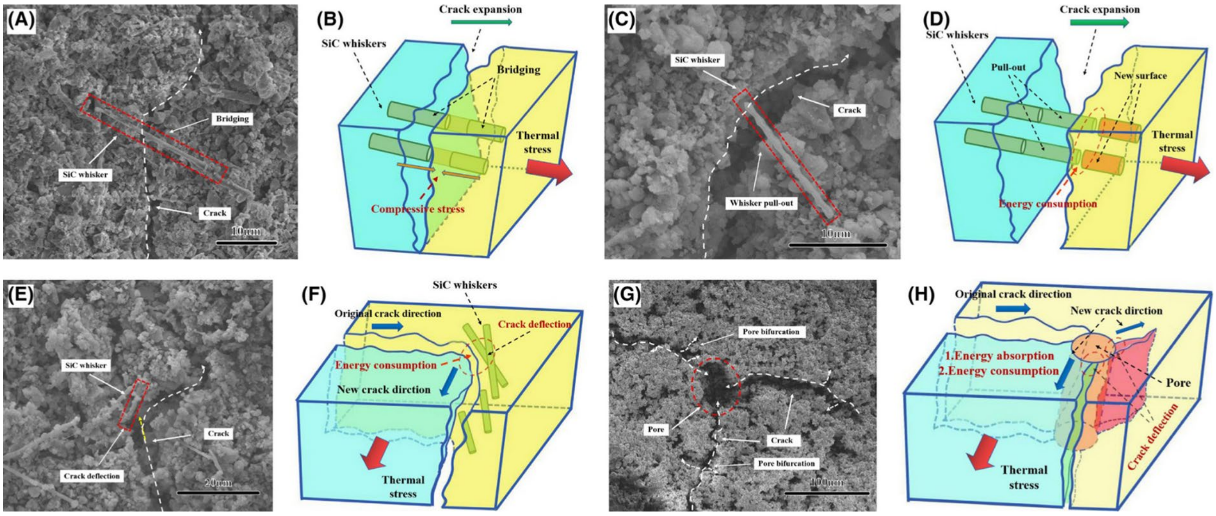


Fig. 3 Representative SEM images and schematic diagrams of c-$AlPO_4$−SiC_w−mullite/SiC coating after 24 thermal cycles (Reproduced with permission from Ref. [63])

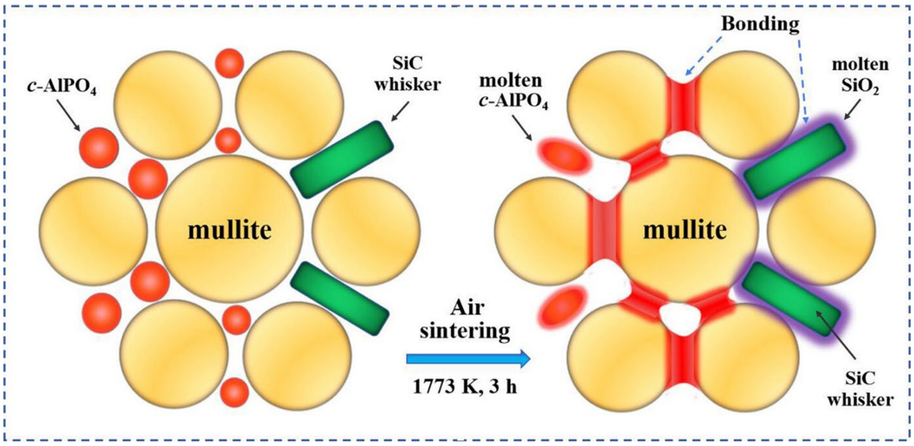


Fig. 4 Schematic diagram of c-AlPO$_4$ and SiC whisker improving bonding strength of c-AlPO$_4$−SiC$_w$−mullite coating after sintering at 1500 °C for 3 h in air (Reproduced with permission from Ref. [63])

the thermal stress, suppressing crack propagation, improving bonding strength and reducing instance of coating cracking or delamination. The incorporation of SiC nanowires markedly improves interfacial bonding strength. The bridging and deflection mechanisms provided by the nanowires mitigate the thermal stress induced by volume changes in the coating, inhibit crack formation and propagation, alter the trajectory of crack propagation, enhance interlayer toughness, and thus improve the thermal cycling stability and mechanical performance of the coating [21,71−73].

3.2 Rare earth silicate modified coating

Rare earth silicates are recognized as excellent materials for thermal barrier coatings; however, they exhibit a different thermal expansion coefficient compared to the C/SiC matrix. To ensure the maintenance of bonding strength between these two components, SiC is necessary as an intermediary bonding layer. Typically, silicates are synthesized in powder form through sol−gel methods, after which rare earth silicate coatings can be applied using techniques such as low-pressure plasma spraying, air spraying, or slurry impregnation methods.

3.2.1 Yttrium silicate modified coatings

A layer of SiC coating can be deposited on the substrate using CVD. Subsequently, a Y_2SiO_5−$Y_2Si_2O_7$ coating is prepared on the SiC layer. Due to the poor adhesion of pure Y_2SiO_5 coatings to the C/SiC composites, a combination of atmospheric plasma spraying (APS) and sol−gel techniques is employed to fabricate yttrium silicate coatings. This approach proves to be significantly more effective than utilizing APS alone for preparing sealing coatings. A high content of $Y_2Si_2O_7$ is essential for ensuring both the thermal expansion coefficient and stability of the coating.

During the stepwise oxidation process at temperatures ranging from 400 to 700 °C, small cracks may develop; however, within a temperature range from 400 to 1600 °C, these cracks can be self-healed and closed through the intermediate layer, thereby enhancing the oxidation protection effect. Experimental results obtained from plasma wind tunnel tests conducted at temperatures between 1350 and 1650 °C indicate that the coating remains intact until reaching 1650 °C. Comparative cyclic oxidation experiments performed on samples at temperatures of 1300 and 1500 °C show that the yttrium silicate coatings prepared using various methods exhibit good oxidation and erosion resistance for the C/SiC matrix [74−76].

3.2.2 Ytterbium silicate modified coating

A SiC adhesion layer was fabricated on C/SiC composites using pulsed CVD, followed by the application of Yb_2SiO_5 or $Yb_2Si_2O_7$ layers through air spray sol−gel techniques, resulting in a composite coating composed of ytterbium silicate and SiC. During a 100 h air oxidation process at 1500 °C, the ytterbium silicate layer was progressively densified, effectively closing the oxygen diffusion pathways within the composite coating and thereby preserving its oxidation resistance. However, residual silicon overflowed from the substrate due to a combination of volumetric expansion forces, capillary forces, gas driving forces, and gravitational effects. This phenomenon led to the formation of new microcracks that compromised the oxidation resistance of the coating [77−79]. The schematic diagrams are shown in Fig. 5 [79].

After the removal of residual silicon, the coating demonstrated excellent oxidation resistance during oxidation processes at temperatures of 1400 and 1500 °C. Initially, good oxidation resistance was observed during high-speed gas thermal shock experiments. However, as time progressed, through-thickness cracks emerged following a 30 min thermal cycle at 1500 °C, which facilitated oxygen diffusion into the substrate and resulted in rapid substrate oxidation. After completing 36 cycles (540 min) of thermal cycling, the outer surface of the coating exhibited significant circular corrosion pits, with the centers of these pits increasingly prone to crack initiation [80].

Based on a two-layer coating, a three-layer composite coating was developed, comprising $La_2Zr_2O_7/Yb_2Si_2O_7/SiC$. This advancement involved the incorporation of lanthanum zirconate ($La_2Zr_2O_7$) into the outer layer. Despite the thermal expansion coefficient mismatch between lanthanum zirconate and ytterbium silicate resulting in numerous cracks, the three-layer coating demonstrated superior oxidation resistance compared to the two-layer coating after 24 cycles (360 min) of thermal cycling at 1500 °C. However, enhancing interlayer bonding strength remains a critical consideration [81].

Another approach involved integrating mullite into the middle layer to create a three-layer $Yb_2Si_2O_7/Al_6Si_2O_{13}/SiC$ coating. At 1500 °C, as oxidation progressed, $Yb_2Si_2O_7$ particles diffused into $Al_6Si_2O_{13}$ layer, forming an eutectic interphase that ultimately compromised the protective capability of the coating and led to its failure [82]. However, at 1400 °C, the densification of the middle and top layers, along with reduced defects, improved the overall oxidation resistance, outperforming the dual-layer $Al_6Si_2O_{13}/SiC$ and

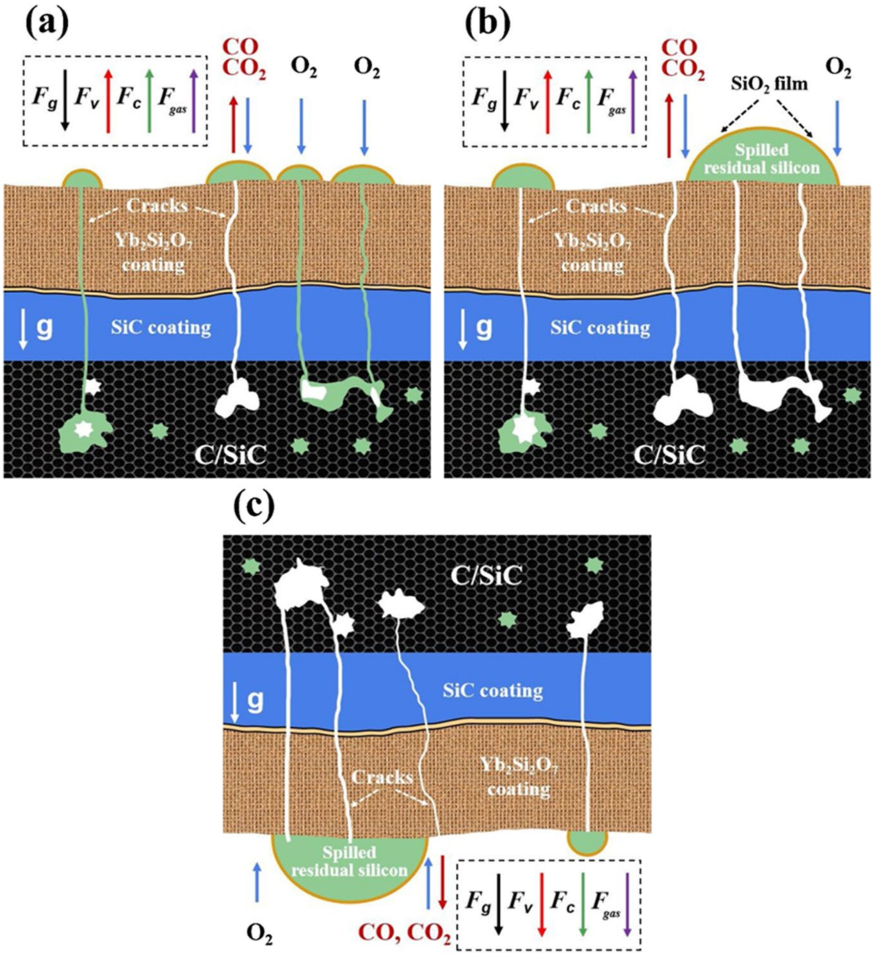


Fig. 5 Schematic diagrams of force on liquid residual silicon with overflowing by bi-layer $Yb_2Si_2O_7/SiC$ coating from C/SiC (B−R−Si) at 1500 °C (Reproduced with permission from Ref. [79])

$Yb_2Si_2O_7$/SiC coatings. The cross-sectional SEM images of three-layer $Yb_2Si_2O_7/Al_6Si_2O_{13}$/SiC coated C/SiC samples after oxidation at 1400 °C for 150 h are shown in Fig. 6 [82]. The three-layer coating minimized the occurrence of through-thickness cracks, diverting the cracks to provide protective effects.

Furthermore, the incorporation of SiC whiskers into the middle layer to create a three-layer $Yb_2Si_2O_7/SiC_w$−mullite (SM)/SiC coating was investigated. Following 200 h of steam-oxygen coupled oxidation at 1400 °C, the dual-layer coating displayed through-thickness cracking, primarily attributed to discrepancies in the thermal expansion coefficients between the inner and outer coatings. However, the addition of SiC whiskers in the three-layer coating significantly improved the resistance of the mullite layer to crack propagation and mitigated the effects arising from differences in thermal expansion coefficients between the outer $Yb_2Si_2O_7$ layer and the middle mullite layer. The SiC whiskers not only alleviated these thermal expansion coefficient disparities but also facilitated silicon dioxide generation during oxidation. This silicon dioxide could react with mullite particles to form a self-healing aluminum silicate glass phase, thereby enhancing the self-healing capability of the mullite coating. Ultimately, this resulted in a markedly improved oxidation resistance for the three-layer coating [68,69,82].

3.3 Boride modified coatings

3.3.1 B_4C modified coatings

As previously mentioned, CVD SiC is the preferred material for repairing surface mechanical cracks and maintaining high residual strength in oxidizing environments. Building on this premise, a self-healing SiC/B_4C dual-layer coating was developed through CVD. In comparison to C/SiC coatings, the incorporation of B_4C within the coating significantly enhances its self-healing capabilities at temperatures below 1000 °C, thereby effectively safeguarding the composite material against oxidation. This enhancement leads to a reduction in coating crack density and a relatively high retention of residual strength. Although this

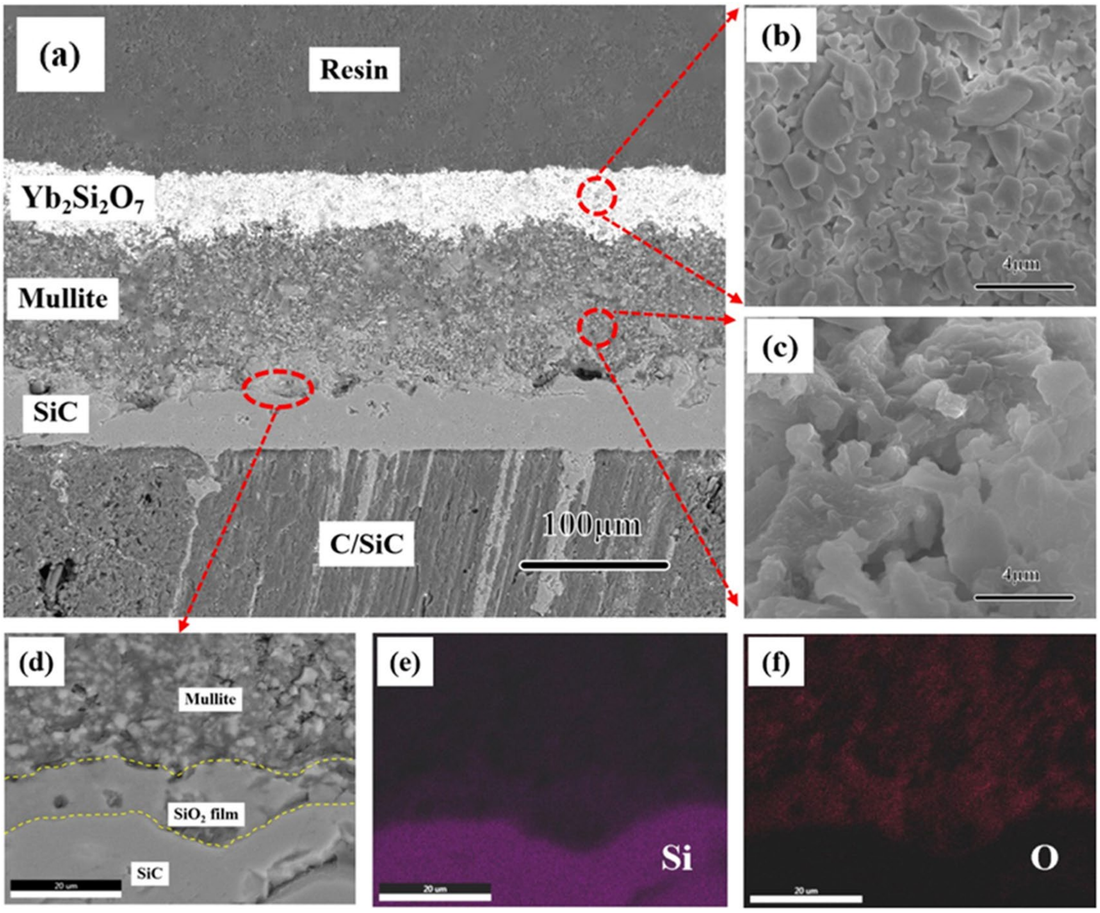


Fig. 6 Cross-sectional SEM images of three-layer $Yb_2Si_2O_7/Al_6Si_2O_{13}$/SiC coated C/SiC samples after oxidation at 1400 °C for 150 h (Reproduced with permission from Ref. [82])

new coating demonstrates lower adhesion strength, it alleviates thermal stress and hinders the transfer of thermal stress, offering superior protection compared to C/SiC coatings. However, it should be noted that beyond 1200 °C, the residual strength of the new coating becomes lower [48,83].

3.3.2 BC_x modified coatings

Composite coatings of $SiC/BC_x/SiC$ and three-layer SiC coatings were fabricated using the CVD method, and their oxidation behavior at 1300 °C was investigated. Although a slower deposition rate can mitigate defects between layers, the former coating demonstrated superior performance in repairing cracks and other defects after being subjected to 1300 °C for 15 h compared to the latter, thereby providing enhanced oxidation protection [30,31,83].

Two multi-layer self-healing coatings, namely SiC/B/SiC and $SiC/BC_x/SiC$, were developed by incorporating boron crystals and boron carbide between two layers of SiC coatings, thereby forming a sandwich-structured composite coating. At 1000 °C, the oxidation of boron crystals and boron carbide results in the formation of boron oxide or $B_2O_3{\cdot}xSiO_2$ films. These films effectively cover cracks and regulate the oxidation process through the diffusion of microcracks along with the presence of glassy boron oxide films. The inclusion of boron carbide significantly reduces mass loss during oxidation, leading to the enhanced strength retention after 10 h of exposure to oxidative conditions. However, at temperatures exceeding 1000 °C, a substantial portion of the film material volatilizes, which diminishes its protective efficacy and adversely affects oxidation resistance [84−86].

The α-BC coating was introduced, and four composite coatings with varying coating ratios were prepared to evaluate the flexural strength and oxidation resistance of the materials after exposure to oxidation at temperatures of 700, 1000, and 1200 °C for a duration of 100 h [87−89].

The SiC/SiC/SiC-coated C/SiC composites (SSS samples) demonstrate low residual strength below 1000 °C, attributed to inhomogeneous oxidation that leads to a reduction in residual strength. Above 1000 °C, many micron-sized defects present in the SSS specimens were repaired by silicon oxide, resulting in enhanced residual strength. However, new defects generated by the volatilization of $Si(OH)_4$ were observed at 1200 °C, which led to another reduction in residual strength. In contrast, the $SiC/\alpha\text{-}BC_x/SiC$ coated composites (SBS samples) exhibited high residual strengths as the oxidation temperature increased at lower annealing temperatures. The coating and composites underwent more significant oxidation, leading to the formation of additional glassy phases with superior oxidation stability compared to those of SSS sample. Above 1000 °C, the cracks in the coatings were completely closed, and the composites with modified coatings demonstrated higher strength retention. This observation confirms that the modified coatings provide effective antioxidant protection for 3D C/SiC composites. However, after holding at 1200 °C for 100 h, the volatilization of $H{-}B{-}O/Si(OH)_4$ and regeneration of coating defects resulted in a slight decrease in coating strength. Notably, the strength is more sensitive to the volatilization of H−B−O compared to $Si(OH)_4$, which leads to inferior residual strength relative to SSS samples.

The $\alpha\text{-}BC_x/\alpha\text{-}BC_x/SiC$-coated composites (BBS samples) exhibit higher residual strength below 1000 °C. However, above this temperature threshold, BBS samples still exhibit excellent residual strength. This finding indicates that BBS samples possess better oxidation resistance within the temperature range from 700 to 1200 °C.

The composite coated with three layers of $\alpha\text{-}BC_x$ coating (BBB sample) demonstrates higher residual strength at temperatures below 1000 °C; however, it exhibits lower residual strength at temperatures exceeding 1000 °C. The residual strength of the coatings is influenced by the oxidation temperature and structural characteristics of the coating.

SBS sample was modified to $\alpha\text{-}BC_x$ by altering the outermost layer to create an $\alpha\text{-}BC_x/\alpha\text{-}BC_x/SiC$ (BBS) coating, which was then compared with the SSS and BBB coatings. Below 1000 °C, the confinement of glassy substances is not significant, resulting in a markedly lower residual strength for the SSS sample compared to the coated samples that incorporate $\alpha\text{-}BC_x$ (BBB and BBS samples). Above 1000 °C, the coated samples containing SiC demonstrate higher residual strengths; conversely, the purely coated $\alpha\text{-}BC_x$ samples (BBB) exhibit reduced strengths. The rational design of $\alpha\text{-}BC_x/\alpha\text{-}BC_x/SiC$ (BBS) provides excellent oxidation resistance across a broad temperature range. The oxidation temperature plays a critical role in determining the self-sealing behavior of BBS coatings, at 700 °C, there is no evident glass sealing effect; however, the oxidation of the α-BC$_x$ layer can consume oxygen and moisture present in cracks, thereby mitigating damage to carbon fibers. Beyond 1000 °C, these cracks can be effectively sealed by the glassy substance [85,87−90]. The microstructural evolution mechanisms of SiC/SiC/SiC (SSS) and SiC/α-BC$_x$/SiC (SBS) annealed at the elevated temperatures are shown in Fig. 7 [90].

3.3.3 Si−B−C modified coatings

The Si−B−C coatings, synthesized by combining Si−B−C with SiC through CVD, resulted in two distinct composite coatings: Si−B−C/Si−B−C/SiC and SiC/Si−B−C/SiC. Oxidation tests conducted over a duration of 50 h in static air at various temperatures indicated that 1000 °C serves as a critical temperature threshold. The incorporation of the Si−B−C layer facilitated the formation of self-healing borosilicate glassy layers, which significantly enhanced oxidation resistance within the temperature range of 700−1300 °C.

Above 1000 °C, the presence of a Si−B−C layer in the intermediate layer allowed the borosilicate glassy layer to effectively seal cracks within the material, thereby significantly enhancing the high-temperature oxidation resistance of the samples. Below 1000 °C, when a double Si−B−C layer was present, the formation of the borosilicate glassy layer was facilitated, resulting in improved self-repairing effects on C/SiC composites and enhanced oxidation resistance.

Under oxidation for 50 h in a moist oxygen atmosphere, the SiC/Si−B−C/SiC coating exhibited

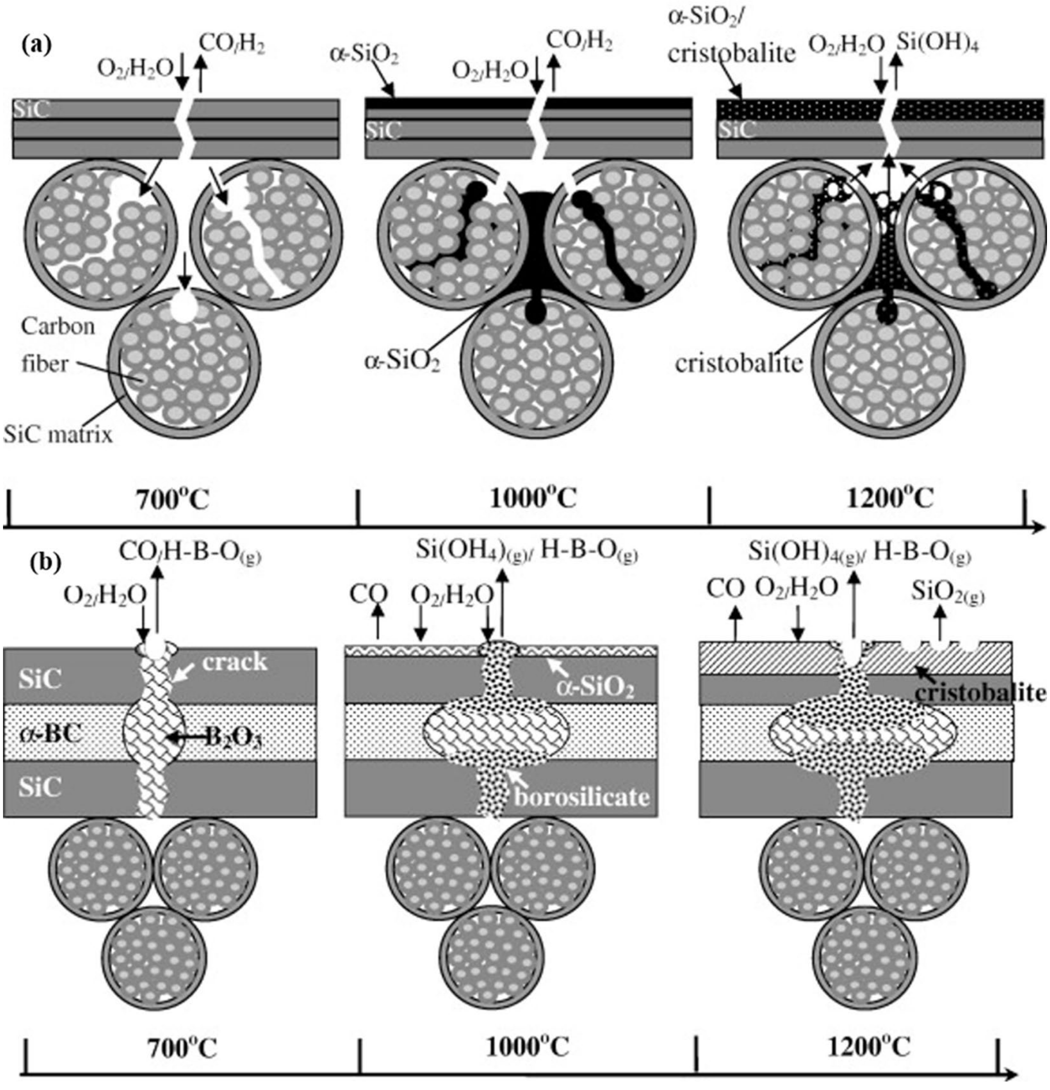


Fig. 7 Microstructural evolution mechanisms of SiC/SiC/SiC (SSS) (a) and SiC/α-BC$_x$/SiC (SBS) (b) annealed in wet oxygen at elevated temperatures (Reproduced with permission from Ref. [90])

high residual strength at 1200 °C. However, temperatures exceeding 1200 °C facilitated the volatilization of borosilicate glassy substances due to increased water vapor pressure. This phenomenon weakened the sealing efficacy of these glassy materials and caused damage to adjacent carbon fibers, resulting in a reduction of residual strength, although it remained above 80%. The formation of borosilicate glassy substances, primarily occurring at cracks and interphase gaps with other coatings, demonstrated an effective approach for sealing cracks and enhancing oxidation resistance in the specific temperature ranges [91,92].

The preparation of multilayer interphase coatings, specifically at PyC/SiC molar ratio n=4, was conducted. Subsequently, dual-layer coatings of SiC/Si−B−C or Si−B−C/SiC were applied onto these interphase coatings. The carbon fiber-reinforced ceramic matrix composites featuring Si−B−C sealing layers demonstrated exceptional high-temperature tensile properties, with values ranging from 219 to 279 MPa. At elevated temperatures from 1200 and 1500 °C, the tensile strength of the coated composite materials exhibited a significant enhancement compared to that of the uncoated composite materials, resulting in an increase of approximately 28%. The carbon fiber-reinforced ceramic matrix composites with SiC sealing layers exhibited a tensile strength of (187±2) MPa at 27 °C. At elevated temperatures of 1000 and 1250 °C, the tensile strength was increased by approximately 18% and 22%, respectively, indicating enhanced mechanical properties at high temperatures. This improvement can be attributed to the multilayer interphases as well as the Si−B−C layer itself, which forms borosilicate glassy phases under high-temperature conditions. These glassy layers progressively mitigate thermal stress and enhance fiber strength, providing protection against oxidation until failure occurs through self-healing borosilicate layers. Consequently, this mechanism prevents fiber debonding and mitigates matrix cracking induced by fiber pullout [93,94].

3.4 Ultra-high temperature ceramic phase (ZrC and ZrB_2) modified coatings

3.4.1 ZrC modified coatings

SiC and ultra-high temperature ceramics, such as zirconium carbide (ZrC) and zirconium boride (ZrB_2), demonstrate remarkable resistance to erosion and oxidation when they were utilized in coatings, particularly in comparison to uncoated C/SiC composites. The integration of these two materials to develop composite coatings with enhanced overall performance represents a promising strategy.

Ablation morphology of SiC-coated samples and ZrC-modified SiC-coated samples after ablation tests, is shown in Fig. 8 [95]. ZrC coatings were synthesized through a combination of chemical vapor phase precipitation and slurry coating methods. Upon oxidation, ZrC transforms into zirconium oxide (ZrO_2), which generates gas and dissipates significant amounts of heat. Concurrently, the resulting ZrO_2 further reacts with carbon during the ablation process, thereby facilitating the oxidation of carbon fibers. Subsequently, ZrC coatings are applied to carbon fibers using the molten salt method to enhance their oxidation resistance. Thermal shock experiments confirmed that C/SiC composites demonstrated superior mechanical properties when protected by the ZrC coating [20,95,96].

The composite coating, featuring tungsten as the intermediate adhesion layer and ZrC as the ablation-resistant layer, was fabricated through plasma spraying. Under conditions of ablation distances of 100 mm for 60 s and of 30 mm for 300 s, the results demonstrated that the composite coating exhibited excellent ablation resistance. The outer ZrC layer plays a crucial role in this performance. During both long-distance short-time and short-distance long-time ablation processes, ZrC undergoes oxidation to form zirconia, resulting in significant gas release that effectively dissipates heat. Subsequently, the molten WO_3 within the intermediate layer also absorbs substantial amounts of heat while filling coating defects to create a dense protective layer over the substrate. In prolonged ablation scenarios at shorter distances, due to temperature gradients, three distinct structural zones emerge from the center to the edge: a central dense zone, a transition zone, and an edge loose zone [97,98].

The composite coatings formulated with SiC and ZrC demonstrate significant brittle fracture behavior when subjected to thermal shock and thermal cycling. These coatings can enhance the

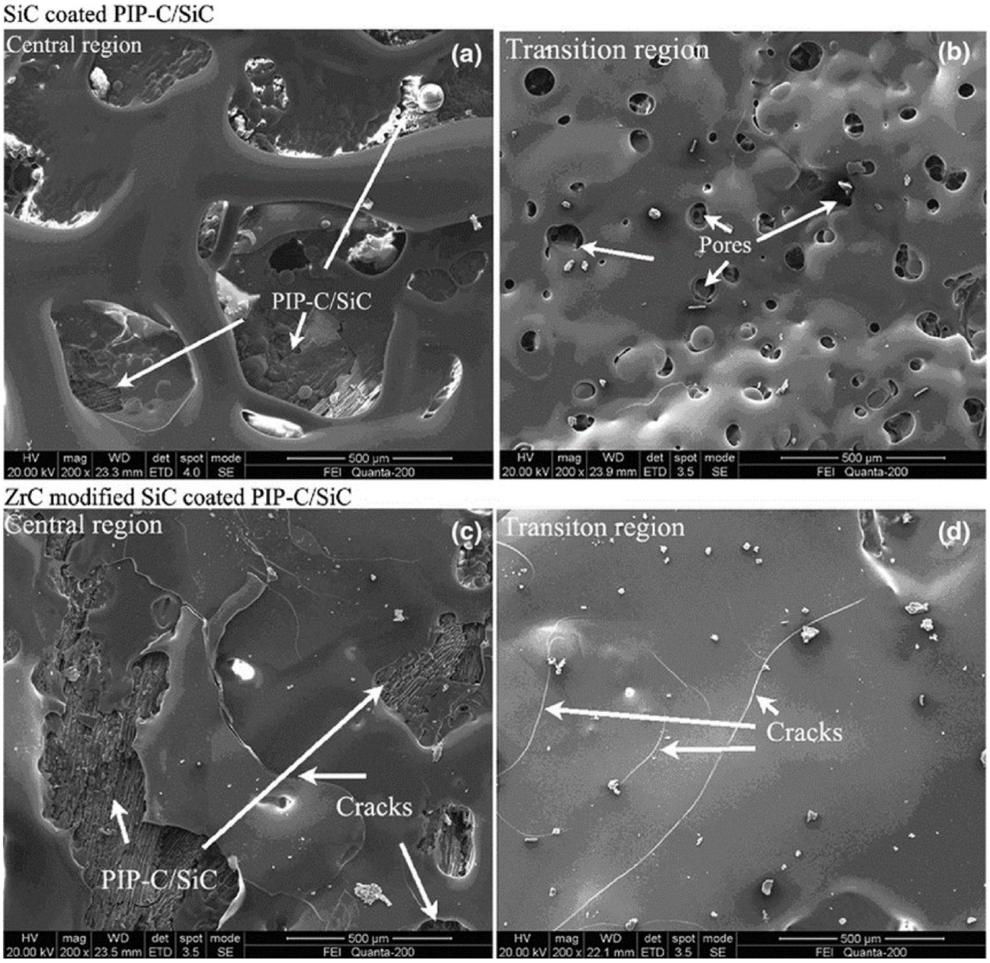


Fig. 8 Ablation morphologies of samples after ablation tests: (a) Central region of SiC-coated samples; (b) Transition region of SiC-coated samples; (c) Central region of ZrC-modified SiC-coated samples; (d) Transition region of ZrC-modified SiC-coated samples (Reproduced with permission from Ref. [95])

mechanical properties of carbon ceramics while preserving their flexural strength. Following exposure to thermal shock, the composite coatings maintain the presence of ZrC, SiC, and carbon [99]. In comparison to single SiC coatings, multilayer coatings comprising SiC and ZrC exhibit remarkable oxidation resistance at elevated temperatures exceeding 1400 °C. This enhanced performance can be attributed primarily to the incorporation of ZrC, which refines the structural integrity of the SiC coating. During oxidation processes, a dense and continuous silica film is formed, effectively impeding oxygen diffusion at high temperatures. Additionally, the formation of a zirconia layer and the deflection of micro-cracks within the composite materials further enhance the oxidation resistance of the coatings [14,96,100,101].

The SiC/ZrC composite coating, in comparison to uncoated carbon fiber-reinforced ceramic matrix composites, demonstrates a denser surface and a significantly reduced ablation rate. The effectiveness of substrate material protection increases with the increase of the thickness of the coating. During the ablation process, a substantial amount of gas is released, which dissipates considerable heat. The ZrO_2-SiO_2 glassy layer formed from the oxidation of Zr−Si−C exhibits a dense surface and exceptional ablation resistance [20,96,102,103].

3.4.2 ZrB_2 modified coatings

ZrB_2 has exceptional oxidation resistance [104,105]. The composite coating formulated with SiC and ZrB_2 exhibits pronounced brittle fracture behavior when subjected to thermal shock and thermal cycling (thermal fatigue). While the flexural strength of carbon ceramics experiences minimal reduction, their mechanical properties are enhanced. Following thermal fatigue, the residual components present in the composite coating consist of ZrB_2, SiC, and carbon. The macrography of C/SiC composites after ablation is shown in Fig. 9 [106]. In comparison to uncoated samples, those with a composite coating demonstrate significant advantages regarding flexural strength retention, ablation rate, and thermal shock resistance. During thermal cycling, the observed reduction in flexural strength can primarily be attributed to fiber−matrix interphase debonding as well as the oxidation of the coated samples [99,106−108].

Under oxy-acetylene flame ablation, an optimal proportion of materials can significantly enhance interfacial bonding strength and coating stability. The oxidation of ZrB_2 and SiC results in the formation of a ZrO_2−SiO_2 layer, which, upon evaporation at elevated temperatures, absorbs heat from the flame. This mechanism effectively reduces erosion of both the carbon fiber and SiC matrix. This process creates a four-layer thermal barrier: an inner SiO_2 layer, a thin discontinuous porous SiO_2 layer, a liquid silicon oxide-sealed porous zirconia layer, and a dense ZrO_2−SiO_2 outer layer. The cross-sectional schematic diagram of the ZrB_2-coated C/SiC composite after ablation is shown in Fig. 10 [109]. The topmost ZrO_2 layer provides essential ablation protection for the underlying coatings. The ablation performance of the composite coating does not exhibit a linear relationship with coating thickness. Instead, as the thickness of the coating increases, residual thermal stress arising from the disparity in thermal expansion coefficients between the coating and substrate can lead to cracks and defects. This phenomenon initially enhances and subsequently diminishes ablation resistance. To improve cyclic ablation resistance, a pre-oxidation treatment of the layered ZrB_2 and SiC coating at 1600 °C is recommended to enhance interfacial structure. During heating, the SiC coating generates a substantial amount of silica glass film, which improves interfacial bonding between SiC and ZrB_2. This enhancement effectively resists oxygen diffusion and mechanical erosion, significantly increasing both the bonding strength of the coating and its cyclic ablation resistance [106,108−114].

Moreover, SiC, ZrB_2, and various other materials can be utilized to formulate multi-component composite coatings.

(1) Addition of $MoSi_2$

A composite coating composed of SiC, ZrB_2, and molybdenum disilicide can be fabricated. Utilizing slurry sintering and atmospheric plasma spraying techniques, a composite coating is prepared using an inner layer of ZrB_2−SiC−glass and an outer layer of ZrB_2−$MoSi_2$−glass. With a glass content of 30% in the inner layer and sintering conducted at 1300 °C for 2 h, the resulting coating attains high density. Ablation test results prove that the coating exhibits excellent ablation resistance at temperatures up to 1500 °C. A dense oxide layer forms on the surface of the coating, consisting of a SiO_2 glass filling phase supported by ZrO_2 and $ZrSiO_4$ phases. The low oxygen diffusion coefficient of SiO_2 ensures that this dense oxide layer effectively functions as an oxygen barrier, thereby inhibiting continuous oxidation of the coating. However, when the temperature reaches 1800 °C, $MoSi_2$ completely volatilizes while $ZrSiO_4$ decomposes into ZrO_2, ultimately leading to failure of the coating [115,116].

Using atmospheric plasma spraying technology, a composite coating was developed using Mo as a transition layer and ZrB_2−SiC as the primary components to investigate the ablation behavior of the composite material. The findings indicate that

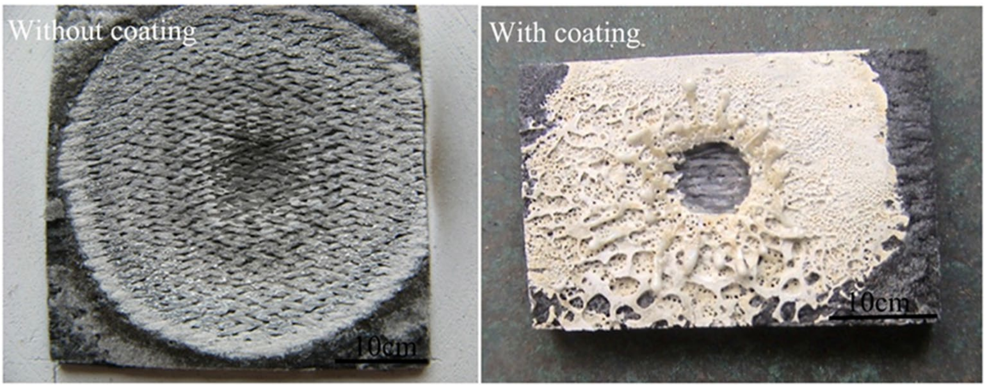


Fig. 9 Macrography of C/SiC composites after ablation (Reproduced with permission from Ref. [106])

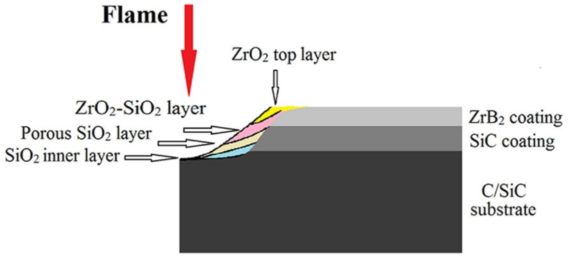


Fig. 10 Cross-sectional schematic diagram of ZrB_2-coated C/SiC composite after ablation (Reproduced with permission from Ref. [109])

the ablation mechanism of this composite is governed by phase transformation, thermal reaction, and thermal diffusion processes. Notably, the solid−liquid transition of ZrB_2 and $MoSi_2$ emerges as the predominant factor influencing ablation performance. The schematic diagrams of reaction progress model and ablation mechanism model are shown in Fig. 11 [117]. Both the thermal reactivity of the coating and its favorable thermal diffusivity significantly contribute to its overall ablation characteristics. Upon the completion of phase transition, it is observed that no continuous liquid phase forms within the composite coating. Ultimately, when all liquid phases are consumed, failure occurs in the coating leading to a rapid degradation of the C/SiC substrate [114,117].

(2) Addition of aluminum nitride

By incorporating aluminum nitride (AlN), a composite coating of ZrB_2−SiC−AlN is formed. This coating exhibits high density, free-cracks, and strong adhesion to the substrate. Following thermal cycling, the primary constituents of the coating are identified as zirconia (ZrO_2) and mullite (a solid solution of Al_2O_3 and SiO_2). The mullite solid solution adheres densely to the surface of the zirconia grains in the form of an oxide film, which remains thermodynamically stable at temperatures up to 1900 °C. The surface film transitions from silica (SiO_2) to a SiO_2−Al_2O_3 solid solution, effectively covering the zirconia composition and collectively enhancing both oxidation resistance and high-temperature endurance of the coating [118,119].

(3) Addition of solid solution elements

A combination of SiC and borides was employed to develop a high-temperature-resistant coating featuring a quaternary boride solid solution of $(Hf_{0.4}Zr_{0.4}Ta_{0.2})B_2$−SiC−Si (HZTS). The results indicate that the coating possesses a dense structure, uniform composition, and high interfacial bonding strength with the substrate, measured at 18.25 MPa. Additionally, the coating demonstrates excellent

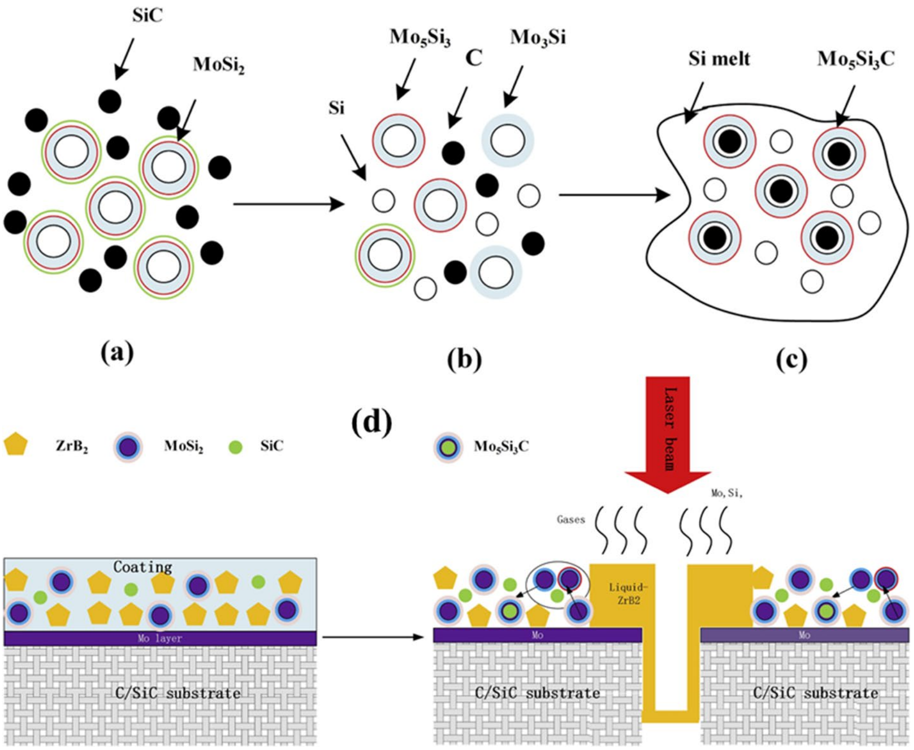


Fig. 11 Schematic diagrams of reaction progress model (a−c) and ablation mechanism model (d) (Reproduced with permission from Ref. [117])

cyclic ablation resistance. The superior ablation resistance is primarily attributed to the formation of a Hf−Zr−Ta−O solid solution skeleton along with a multi-component silicate glass composite structure. This configuration significantly impedes oxygen diffusion into the HZTS coating and enhances its ability to withstand mechanical erosion caused by airflow impact, and the synergistic effect of the outer oxidation layer combined with the inner dense HZTS coating ensures long-term ablation resistance for C/SiC composites [120].

3.4.3 Ultra-high temperature ceramic phase (ZrC and ZrB_2) co-modified coatings

In studies focused on ZrB_2 and SiC coatings, the incorporation of ZrC represents a significant area of research. The addition of ZrC effectively enhances ablation resistance by forming a dense and complete ZrO_2−SiO_2 layer, which substantially mitigates oxygen infiltration. The design of the coating comprises an inner layer composed of C/SiC composites and an outer layer consisting of C/SiC−ZrB_2−ZrC. During 1000 s cyclic ablation tests, the initial two cycles of the C/SiC−ZrB_2−ZrC coating reveal that ZrB_2, ZrC, and SiC undergo oxidation to form ZrO_2, B_2O_3, and SiO_2. In the subsequent four cycles, a significant evaporation of glassy oxides is observed. Within a single ablation cycle, the process begins with the oxidation of ZrB_2, ZrC, and SiC into their respective oxides (ZrO_2, B_2O_3, and SiO_2), which is then followed by mechanical erosion of these oxidation products. This alternates with the oxidation of carbon fibers present in the preforms. Despite the potential reduction in flexural strength due to possible delamination at the coating−substrate interphase, both SiC and ultra-high temperature ceramics (such as ZrC and ZrB_2) coatings can significantly enhance the mechanical properties of ceramics. At varying temperatures, ZrB_2 and SiC coatings demonstrate superior oxidation resistance compared to ZrC and SiC coatings, thereby effectively protecting ceramics from oxidative degradation [14,113,119,121−123].

4 Conclusions and outlook

C/SiC materials demonstrate exceptional mechanical properties, high specific strength, and significant hardness, which effectively contribute to mass reduction in structures and prolong their service life. These materials find extensive applications in aerospace, automotive manufacturing, and energy sectors. However, despite the superior performance of C/SiC materials, components may experience gradual failure under prolonged exposure to high-temperature oxidation conditions due to inadequate bonding strength between the coating and substrate. This can lead to delamination and crack propagation, thereby limiting their broader applicability. The application of SiC and its modified coatings can enhance the overall performance of C/SiC materials. This study classifies SiC and its modified coatings and by combining the macroscopic performance, microstructure, stress analysis, and elemental distribution of various coating samples, the evolution of mechanical and oxidation mechanisms of SiC and its modified coatings is analyzed, providing reliable reference for the future development of new SiC and its modified coatings.

When a SiC interphase replaces pyrolytic carbon deposited on carbon fibers, it generates self-healing glassy silica at specific temperatures, thereby providing protection to the carbon fibers. However, as the temperature increases and cracks develop along with the formation of oxygen channels, the coating undergoes rapid oxidation. Although depositing SiC on the surface of C/SiC composites offers advantages such as high-temperature resistance, oxidation resistance, and enhanced strength, it encounters the accelerated oxidation at certain elevated temperatures. Furthermore, challenges related to insufficient bonding strength between the SiC coating and its substrate, as well as issues concerning high-temperature oxidation, necessitate modifications to the coating for the improved performance.

Modifications of SiC-based coatings can be categorized into structural and compositional modifications. Structural modifications involve the incorporation of SiC whiskers and nanowires, which enhance the bonding strength between the coating and substrate, mitigate thermal stress within the coating, inhibit crack initiation and propagation, and improve both thermal cycling resistance and mechanical properties of the coating. Notably, composite coatings that consist of whiskers, c-$AlPO_4$, and mullite demonstrate significant improvements in thermal shock resistance as well as self-healing capabilities.

Compositional modifications can be classified into three categories: (1) Coatings modified with rare earth silicates enhance the oxidation resistance and thermal shock performance of the coating samples. During the oxidation process, rare earth silicates continuously repair the oxygen diffusion channels that are formed due to the elevated temperatures, thereby improving the overall oxidation resistance of the composite material at temperatures exceeding 1400 °C; (2) Boride-modified coatings demonstrate a critical temperature threshold at 1000 °C. The incorporation of borides facilitates the formation of self-healing borosilicate glassy layers ($B_2O_3{\cdot}x SiO_2$ film), which exhibit remarkable oxidation resistance within the temperature range of 700−1200 °C; (3) Ultra-high temperature ceramic phase modified coatings, such as ZrC and ZrB_2, develop a three-dimensional ZrO_2-SiO_2 structural layer during ablation and oxidation processes. This mechanism generates a significant amount of gas that effectively dissipates heat, thereby enhancing the overall oxidation and ablation resistance of the coating material.

Through the integrated application of structural and compositional optimizations, the performance of SiC modified coatings can be significantly enhanced. Future research will be concentrated on two primary areas: Firstly, a more in-depth investigation into the interphase structure between the coating and substrate, examining the effects of various nanomaterials (such as nanoparticles and nanowires) on interfacial bonding strength. This includes the development of buffer interlayers to mitigate the interfacial stress induced by thermal expansion discrepancies between the coating and substrate, thereby enhancing interfacial bonding strength. Secondly, the scientific and rational combination of the modified components (such as borides and rare earth silicates) is essential for developing multifunctional coatings. These coatings should simultaneously enhance oxidation resistance, thermal shock resistance, and ablation resistance, thereby expanding their application potential in high-temperature, high-pressure, and corrosive environments.

CRediT authorship contribution statement

Wu-bin QI: Investigation, Methodology, Writing – Origin draft, Review & editing; **Jia-qui WU:** Suppervision, Writing − Review & eding; **Zhuan LI** and **Peng XIAO:** Investigation, Conceptualization, Funding acquisition, Supervision, Writing − Review & editing; **Jun-jie DUAN**, **Yang-jie LI**, **Liang PANG**, **Zong-long GAO**, **Jia-min ZHU** and **Yang LI:** Supervision, Writing − Review & editing.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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C/SiC 复合材料中碳化硅及其改性涂层的研究进展

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摘 要：碳纤维增强碳化硅陶瓷复合材料(C/SiC)以其优异的耐高温、耐腐蚀和抗氧化等性能成为关键的结构材料，但 C/SiC 复合材料在长期高温载荷条件下易发生氧化。本文综述了 C/SiC 复合材料表面 SiC 涂层及其改性涂层的研究进展，这些涂层的优化主要涉及结构和成分两个方面：结构优化的重点是通过 SiC 纳米线等改善界面相结构，从而提高涂层之间以及涂层与基体之间的界面结合强度。成分优化的重点是通过稀土硅酸盐等成分来改善涂层在高温载荷下的性能等。重点介绍了各种改性涂层的改性策略，有助于指导未来高性能 C/SiC 涂层材料的制备。

关键词：C/SiC 复合材料；涂层；涂层改性；氧化；强度

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