

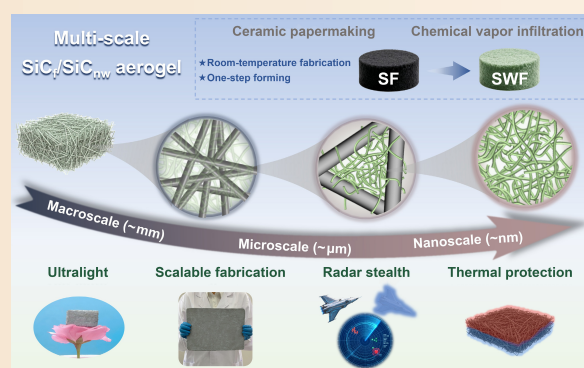
Multiscale SiC fiber/nanowire aerogels for broadband electromagnetic wave absorption via scalable ceramic papermaking coupled with CVI

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ABSTRACT: The escalating demand for multifunctional thermal protection systems (TPSs) necessitates advanced materials that concurrently withstand extreme environments and mitigate electromagnetic wave (EMW) radiation. However, single-scale coarse fiber frameworks often suffer from limited polarization interfaces and loose mechanical interlocking. To address this, by coupling a scalable room-temperature ceramic papermaking strategy with chemical vapor infiltration (CVI), we fabricate multiscale all-ceramic SiC_f/SiC_{nw} aerogels (SFWs) that seamlessly bridge millimeter-scale fibers to nanoscale nanowires. Operating entirely under ambient conditions without the need for size-limiting equipment, this highly scalable and versatile strategy enables the fabrication of large-area components with arbitrary geometries. Optimizing the matrix fiber length to 5 mm unlocks sufficient interfiber void space for the uniform proliferation of ultrahigh-aspect-ratio SiC nanowires. The resulting SFW-5 M aerogel achieves an ultralow density of merely 0.23 g·cm⁻³ alongside an exceptional minimum reflection loss (RL_{min}) of -46.62 dB. Under 1000 °C flame exposure, the back-surface temperature is reliably restricted to approximately 250 °C. Crucially, rather than undergoing catastrophic embrittlement, prolonged ablation triggers a striking “fire-strengthening” effect, simultaneously enhancing compressive peak stresses and expanding the effective absorption bandwidth (EAB) to 11.5 GHz—effectively spanning the C, X, and Ku bands (72% of the 2–18 GHz range). Driven by cross-scale synergy between the fibrous scaffold and nanowire network, these durable aerogels overcome the inherent limitations of single-scale scaffolds, offering a scalable paradigm for next-generation aerospace TPS design.

KEYWORDS: SiC aerogel; multiscale architecture; ceramic papermaking; electromagnetic wave absorption; fire-strengthening



1 Introduction

Rapid aerospace advancements in hypersonic vehicles and reusable spacecraft have created an urgent demand for lightweight [1], multifunctional thermal protection systems (TPSs) capable of concurrently mitigating electromagnetic wave (EMW) absorption [2], blocking intense heat flux [3,4], and retaining mechanical integrity under extreme service environments [5–7]. Aerogels, by virtue of their ultrahigh porosity and ultralow density, have emerged as a primary platform for such applications. To date, various aerogel scaffolds based on carbon [8], MXenes [9], and oxide ceramics [10] have been extensively investigated. However, these conventional candidates often fail to simultaneously satisfy the stringent requirements of extreme thermal stability and sustained functional performance [11]. Carbonaceous aerogels suffer from catastrophic oxidation at elevated temperatures,

whereas oxide ceramics typically exhibit limited EMW attenuation capability due to their insulating nature [12]. Consequently, silicon carbide (SiC) ceramics are universally recognized as one of the most promising material systems for high-temperature multifunctional structures, fundamentally rooted in their high melting point, exceptional oxidation resistance, and highly tunable dielectric properties [13,14].

Despite the immense potential of SiC, the fabrication of ultralight SiC aerogels remains encumbered by significant processing bottlenecks. Conventional sol–gel routes necessitate complex, multistep procedures involving supercritical drying and high-temperature sintering, which are not only energy intensive and cost prohibitive but also intrinsically difficult to scale for large-area or arbitrarily shaped components [15,16]. Although emerging combustion synthesis strategies can dramatically shorten processing timeframes, precisely controlling structural uniformity

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and porosity across macroscopic volumes remains a formidable challenge [17]. Alternatively, the chemical vapor infiltration (CVI) method has been employed to construct SiC aerogels with elasticity conferred by ultrahigh-aspect-ratio nanowires [18,19]. Nevertheless, this single-scale nanowire architecture is highly vulnerable to high-temperature oxidation. Prolonged thermal exposure frequently induces catastrophic embrittlement and structural collapse [20]. Moreover, this reliance on a singular electromagnetic attenuation mechanism, primarily conduction loss within a sparse nanowire network, results in poor EMW absorption performance [21]. Thus, a critical unresolved challenge is to devise a scalable, low-cost fabrication strategy that yields ultralight SiC aerogels while ensuring long-term oxidation resistance, robust mechanical resilience, and stable, broadband EMW absorption after harsh environmental service.

In our previous work, inspired by traditional Chinese papermaking technology, we proposed a “ceramic papermaking” strategy that enables gravity-driven, room-temperature self-assembly of SiC fibers into ultralight aerogels [22]. This ambient-condition process operated entirely without high-temperature or high-pressure steps, eliminating the need for expensive supercritical drying equipment and complex posttreatment protocols. Free from stringent equipment constraints, this method allowed the flexible, cost-effective production of aerogels in arbitrary sizes and geometries while enabling precise control over aerogel density to meet diverse application requirements. Owing to the relatively large diameter of the constituent fiber, these single-scale fibrous aerogels exhibited remarkable thermal stability, withstanding severe exposure to 1600 °C without structural collapse. Remarkably, postablation characterization revealed a distinct “fire-strengthening” phenomenon, wherein the compressive strength increased from 49 to 206 kPa after thermal treatment. Nonetheless, the absence of small-scale constituents within these single-scale scaffolds severely restricts interfacial polarization and conductive network formation, leading to intrinsically limited EMW absorption performance. This performance trade-off underscores a fundamental bottleneck in current SiC research systems. Single-scale architectures fail to decouple mechanical/thermal robustness from electromagnetic wave attenuation, as coarse fibers lack polarization interfaces, while fine nanowires lack macroscopic structural durability. To eliminate the EMW absorption limitations inherent to single-scale fiber networks without compromising their processing scalability, transforming the ceramic papermaking route into a multiscale precursor platform offers distinct architectural advantages. The room-temperature fiber aerogel acts as an open scaffold that inherently guarantees mass production and extreme thermal resistance while providing accessible void spaces for the uninhibited growth of nanoscale dissipation networks.

Herein, to overcome the inherent limitations of single-scale architectures, we report a multiscale all-ceramic SiC_f/SiC_{nw} (SFW) aerogel design that integrates a room-temperature ceramic papermaking process (for macroscaffold assembly) with a subsequent CVI to hierarchically grow nanometer-scale SiC nanowires (SiC_{nw}) within the millimeter-scale SiC fiber (SiC_f) network. In this architecture, the millimeter-scale fiber scaffold, fabricated via ambient-condition gravity settling, provides a lightweight, mechanically robust macroscaffold with exceptional thermal insulation and oxidation resistance. Concurrently, CVI-grown nanowires uniformly infiltrate the interfiber voids, introducing a dense nanoscale network that dramatically increases the specific surface area, generates abundant heterogeneous interfaces for intense interfacial polarization, and establishes

interconnected conductive pathways for enhanced dielectric loss. This cross-scale synergy between the fibrous scaffold and the nanowire filler is systematically modulated by adjusting the matrix fiber length. We demonstrate that optimizing the fiber length to 5 mm unlocks sufficient open space for the uniform proliferation of ultrahigh-aspect-ratio SiC nanowires, yielding a multiscale aerogel (SFW-5 M) with an ultralow density of merely 0.23 g·cm⁻³. The resulting hierarchical structure achieves a RL_{min} of -46.62 dB and restricts the back-surface temperature to approximately 250 °C under a 1000 °C flame test. Crucially, prolonged high-temperature ablation triggers a pronounced fire-strengthening effect, simultaneously boosting compressive peak stresses and expanding the effective absorption bandwidth (EAB) to an exceptional 11.5 GHz—effectively spanning the C, X, and Ku bands. By bridging millimeter to nanometer scales through a low-cost, scalable fabrication paradigm, these durable multiscale aerogels overcome the performance bottlenecks of single-scale scaffolds, providing a practical and highly adaptable reference for next-generation aerospace TPS design.

2 Experimental

2.1 Raw materials

SiC fibers (SLF-II, monofilament diameter 10.0±1.0 μm) were purchased from Suzhou Cerafil Ceramic Fibers Co., Ltd., China. Fumed SiO₂ (S861488, > 99%), SiO₂ (S818287, > 99%), CaCO₃ (C805317, > 99%), and carbon powder (A708556, > 99%) were obtained from Shanghai Macklin Biochemical Technology Co., Ltd., China. Short carbon fibers (C_f, Φ: 5–10 μm, L: 20–50 μm) were supplied by ENO High-tech Materials Development Co., Ltd., China. Cetyltrimethylammonium bromide (CTAB; H108986, ≥ 99.0%) was purchased from Shanghai Aladdin Co., Ltd., China, and used as a cationic dispersant. Triethylamine (M03649, ≥ 99.5%) was sourced from Shanghai Merrier Biochemical Technology Co., Ltd., China. Hydrochloric acid (HCl, CP, 36.0%–38.0%) was provided by Sinopharm Chemical Reagent Co., Ltd., China. Ethanol (EtOH; M34056, ≥ 99%), methyltrimethoxysilane (MTMS; M03116, 98%), and dimethyldimethoxysilane (DMDMS; M02235, 97%) were obtained from Shanghai Meryer Chemical Technology Co., Ltd. (Shanghai, China). Silicon powders (325 mesh, purity ≥ 99.99%) were supplied by Forsman Scientific (Beijing) Co., Ltd., China. All chemicals were used as received without further purification.

2.2 Sample preparation

2.2.1 Ceramic papermaking

The fabrication of SiC fiber aerogels (SFs) was conducted via a room-temperature, gravity-driven ceramic papermaking process, as detailed in our previous report [22]. Briefly, deionized water, CTAB, and triethylamine were sequentially added to a wide-mouth beaker, and the pH value was adjusted to 8.5. Fumed SiO₂ was then introduced under continuous stirring for 30 min to form a uniform suspension. Subsequently, SiC_f with distinct lengths of 1, 3, and 5 mm (Figure S1 in the Electronic Supplementary Material (ESM)) were slowly dispersed into the SiO₂ suspension. Stirring was maintained until a homogeneous, CTAB-stabilized SiC_f suspension was formed. The pH was then lowered to 3.0 by adding HCl, which neutralized the cationic dispersant and triggered the spontaneous sedimentation and self-assembly of the fibers into an interconnected three-dimensional network. Excess SiO₂ was thoroughly removed through repeated acid washing. Finally, the as-formed SFs were dried at 60–100 °C for 24 h to

ensure complete dryness. According to the initial SiC_f length, the resulting samples were designated SF-1 M, SF-3 M, and SF-5 M.

2.2.2 CVI

Prior to CVI, the SFs were thermally pretreated at 900 °C for 1 h in air within a muffle furnace to remove residual carbon and prevent fiber degradation during the subsequent high-temperature synthesis. Concurrently, a siloxane precursor sol was prepared by blending MTMS, DMDMS, EtOH, and deionized water at a ratio of 4 : 1 : 5 : 3. The hydrolysis reaction was completed under continuous stirring for 30 min. To this solution, 5 wt% of short carbon fibers (C_f) was added, and the mixture was stirred for an additional 30 min to yield a homogeneous suspension. The pretreated SFs were fully immersed in the suspension to ensure thorough infiltration, after which excess sol was removed via vacuum filtration. The wet composites were then transferred to an oven and maintained at 100 °C for 5 h to induce cross-linking, yielding the SF/xerogel intermediates (the corresponding chemical pathways are illustrated in Fig. S2 in the ESM).

For CVI synthesis, the SF/xerogel was embedded in a powder bed composed of SiO₂/Si (1 : 1) and CaCO₃/C (1 : 1) within a graphite crucible. The deposition was carried out in a graphite furnace (CX2300S, Anhui Chenxin Weike Industrial Technology Co., Ltd., China) under an argon atmosphere. The temperature was ramped to 1450 °C at a heating rate of 5 °C·min⁻¹ and held for 4 h. To eliminate residual carbon, the samples were subsequently calcined at 500 °C for 30 min in air using a muffle furnace (FMJ-07/11, Hefei Facerom Thermal Equipment Co., Ltd., China). This process yielded multiscale SiC_f/SiC_{nw} all-ceramic aerogels. The resulting samples were labeled SFW-1 M, SFW-3 M, and SFW-5 M, corresponding to their initial SiC_f lengths of 1, 3, and 5 mm, respectively.

2.2.3 Characterization

The phase structures were identified by X-ray diffraction (XRD; XRD-6000 X, Shimadzu, Japan) using Cu K α radiation at a scan speed of 5 (°)·min⁻¹. The bulk density of each specimen was calculated from the ratio of its measured mass to geometric volume. The size of the bulk used for density testing was 100 mm $\times$ 100 mm $\times$ 10 mm. Six sets of parallel tests were prepared for each group of samples. Microstructural morphologies and energy-dispersive spectroscopy (EDS) elemental mappings were examined using a scanning electron microscope (SEM; Zeiss Gemini 360, Carl Zeiss, Germany). Uniaxial quasistatic compression tests were performed at a constant crosshead speed of 1 mm·min⁻¹ using an Instron universal testing machine (M10-17410-CS, Instron, USA) equipped with a 1 kN load cell to evaluate the mechanical responses. The electromagnetic wave absorption characteristics were investigated across a frequency range of 2.0–18.0 GHz using a vector network analyzer (VNA; AV3672D, Ceyear, China). The dimensions of the sample tested by the waveguide method are shown in Table S1 in the ESM. Thermal conductivity was evaluated by the transient plane source (TPS) method on a hot disk analyzer (TPS 2500S, Hot Disk AB, Sweden). The size of the thermal conductivity test sample was 25 mm $\times$ 25 mm $\times$ 5 mm. Three parallel tests were conducted for each group of samples. Prior to testing, specimen surfaces were polished flat and slightly pressed against the sensor to ensure minimized contact resistance. High-temperature flame resistance was evaluated using an adjustable butane torch (Model SG006, FOCUS, China). The size of the sample for the high-temperature test was 100 mm $\times$ 60 mm $\times$ 8 mm. During the flame test, an infrared thermal camera (IR; VariocAM HD Head 1024, InfraTec

GmbH, Germany) and a multichannel data logger (TCP-500X, Measure Fine, China) were simultaneously employed to monitor the real-time surface temperature evolution. Additionally, transmission electron microscopy (TEM; Themis Z, Thermo Scientific, USA) was utilized to examine the postablation morphology and crystal structures of the SiC nanowires.

3 Results and discussion

3.1 Design and fabrication

As a traditional Chinese technique, papermaking achieves the shaping of pulp fibers via a straightforward sedimentation process. Inspired by this concept, our prior work demonstrated that ceramic fibers can undergo gravity-driven, room-temperature self-assembly to form ultralight SFs. Herein, SiC nanowires featuring ultrahigh aspect ratios are subsequently introduced into the interfiber voids via CVI, yielding hierarchically multiscale SiC_f/SiC_{nw} aerogels (SFWs), as schematically illustrated in Fig. 1(a). Spanning from the millimeter-scale fiber scaffold to the microscale intrascaffold pores filled with nanowires and down to the nanoscale nanowires themselves, these hierarchical aerogels exhibit an ultralow density. Simultaneously, benefiting from this extremely simple, equipment-size-independent strategy, SFs and SFWs demonstrate excellent scalability, enabling the fabrication of large-scale aerogels (400 mm $\times$ 300 mm) with arbitrarily complex shapes (Fig. 1(a) and Fig. S3 in the ESM). Concurrently, it maximizes synergistic cooperation between the fibers and nanowires, thereby demonstrating significant advantages for concurrent thermal protection and EMW absorption applications.

The millimeter-scale SF scaffolds are fabricated via a facile, scalable ceramic papermaking strategy. Initially, the electrostatic adsorption of CTAB enables the formation of a stable and homogeneous SiC fiber slurry (Fig. 1(b)). As the pH decreases, the adsorption of H⁺ ions destabilizes the suspension, triggering gravity-driven sedimentation and self-assembly to yield an interconnected three-dimensional network. To achieve the targeted multiscale architecture, SiC_{nw} is subsequently introduced into the interior voids of the SFs via CVI. Prior to CVI, a mild preheat treatment is performed on the SFs to maintain their structural integrity. Specifically, the SFs are impregnated with a siloxane sol, followed by vacuum filtration to remove excess precursor. The resulting cross-linked SF/siloxane xerogel is then subjected to the subsequent CVI process within a graphite furnace.

During heating, the surface-adhering siloxane xerogel transforms into silicon oxycarbide (SiO_xC_y). At 1450 °C, the decomposition of SiO_xC_y and the disproportionation between SiO₂ and Si nanoparticles collectively generate SiO and CO vapors (Reaction (1)). Simultaneously, CaCO₃ decomposes to release CO₂ and CO gases (Reactions (2) and (3)). This vapor influx saturates the graphite crucible, establishing the required gaseous environment for nanowire growth. SiC nucleation proceeds via two concurrent pathways: The reaction of SiO vapor with active carbon sites on the fiber surfaces (Reaction (4)) and the carbothermal reduction of the SiO_xC_y matrix (Reaction (5)). These nuclei subsequently guide the growth of ultralong nanowires through continuous gas-phase reactions between SiO and CO (Reaction (6)), as shown in Fig. 1(b). Ultimately, the dense proliferation of SiC nanowires within the microscale pores of the SF scaffold yields an interconnected, multiscale architecture, successfully achieving an ultralight SiC_f/SiC_{nw} aerogel that integrates macroscopic fibers with a nanoscale nanowire network.

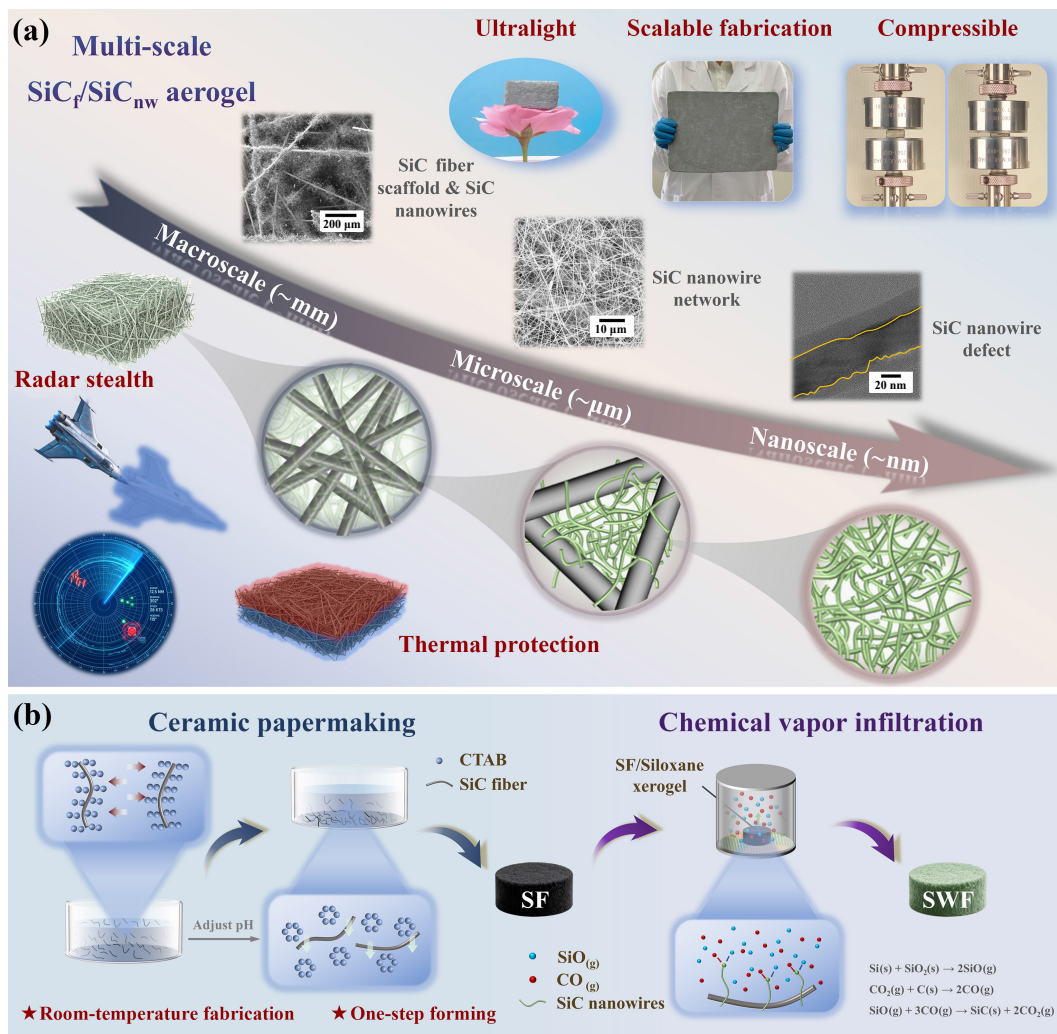
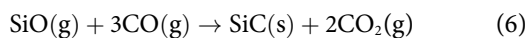
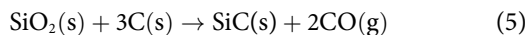
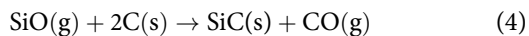
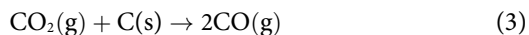
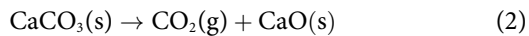
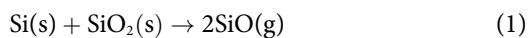


Fig. 1 Schematic diagram of the structural configuration and synthetic route for multiscale all-ceramic aerogels. (a) Multiscale structural organization of the SFW aerogels and their dual-functional utilization in EMW absorption and thermal insulation. (b) Synthetic approach integrating ceramic papermaking of SiC fiber scaffolds with CVI of SiC nanowires.



3.2 Morphologies and mechanical properties

During the ceramic papermaking process, fiber manipulation is achieved solely by modulating the distribution of surface species via electrostatic interactions. Because the fibers undergo no chemical reactions during this procedure, they retain a phase structure identical to that of the pristine fibers, as illustrated in Fig. 2(a). XRD patterns show that both the SFs and the pristine SiC fibers exhibit only a single broad reflection at approximately

35.6°, which directly demonstrates the amorphous nature of the SiC. Notably, although residual silicon oxycarbide (SiOC) phases are present in SFW-1 M, no discrete SiOC diffraction peaks are detected, which is attributed to its highly amorphous structure that primarily contributes to diffuse background scattering rather than sharp Bragg reflections. This observation is highly consistent with the well-documented precursor-derived conversion process used to fabricate SiC fibers.

The fiber length exerts a direct and intuitive influence on the microstructural morphology and density of the SFs. Figure 2(b) and Table S2 in the ESM illustrate the bulk densities of the SFs fabricated using different fiber lengths. Notably, as the fiber length increases from 1 to 5 mm, the density of the SFs decreases from 0.23 to 0.13 g·cm⁻³, which represents only 4% of the density of bulk SiC. According to the mechanical entanglement mechanism governing fibers with an aspect ratio greater than 5, long fibers intertwine to form flocs that trap substantial amounts of air internally while generating large macrovoids between adjacent flocs [23]. This irreversible entanglement effect becomes increasingly pronounced with longer fibers, ultimately driving the reduction in density. This structural variation is further corroborated by the microstructural morphologies presented in Fig. 2(c), which reveal a distinct trend toward lower packing density with increasing fiber length. Compared to SF-1 M, SF-5 M

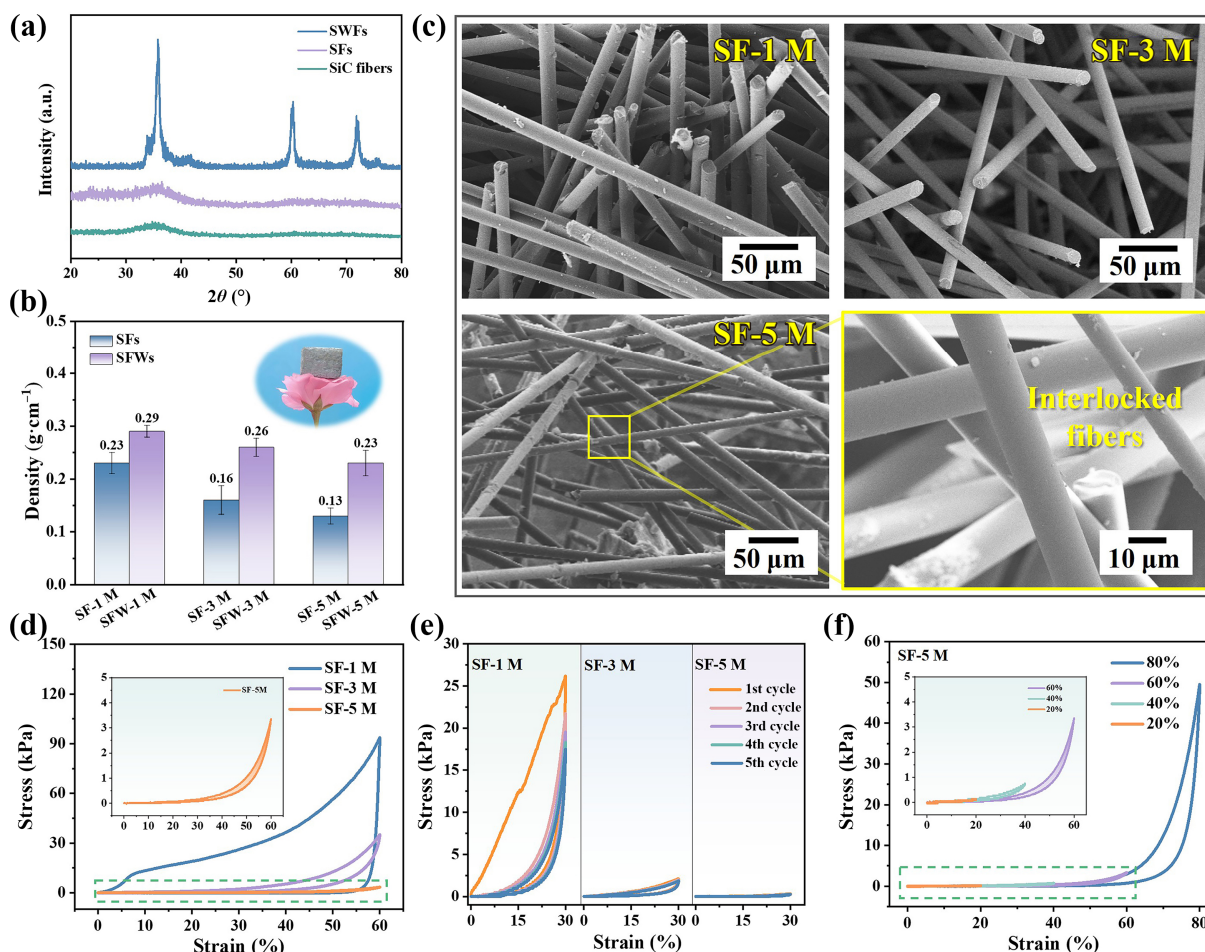


Fig. 2 Structure–property correlations of the SF and SFW aerogel series. (a) XRD profiles of pristine SiC fibers, as-prepared SFs, and SFW aerogels. (b) Bulk densities of SFs and SFWs are plotted against fiber length, alongside the ultralight characteristic of SFW-5 M resting on a flower. (c) SEM micrographs depicting the microstructural evolution. (d) Compressive loading–unloading stress–strain curves evaluated at 60% strain. (e) Cyclic compressive behaviors for five continuous cycles at a preset strain of 30%. (f) Compressive responses of the SF-5 M aerogel under various maximum strains.

exhibits significantly larger interfiber clearance and a more pronounced degree of fiber entanglement. Crucially, the magnified view of SF-5 M clearly captures this mechanical interlocking and fiber wrapping, strongly validating the fiber entanglement theory and the consolidation mechanism of the ceramic papermaking process.

The influence of fiber length on the mechanical properties of ultralight SiC fiber aerogels is further investigated. During compression at 60% strain, the SFs exhibit a characteristic three-stage behavior, as shown in Fig. 2(d). In the initial compression stage, the stress–strain curve ascends slowly, corresponding to the irreversible fiber-sliding process within the aerogel scaffold. Upon completion of this sliding stage, the constituent fibers achieve a fully interlocked state that arrests further slippage. As the load increases, the fibers undergo elastic bending (Fig. S4 in the ESM), which serves as the primary mechanism underlying the excellent elasticity of the SFs. Consequently, the stress escalates more significantly with increasing strain. As compression proceeds, the material enters the fiber compaction stage, where the stress–strain curve steepens sharply. Notably, with increasing fiber length, the peak stress decreases from 93.6 to 3.3 kPa. This reduction occurs because the larger interfiber clearance prevents SF-5 M from reaching a fully compacted state at 60% strain. Concurrently, because longer fibers accommodate a greater degree of deformation, SF-5 M demonstrates superior compressive resilience, as highlighted in the inset of Fig. 2(d). Cyclic compression tests on different samples further substantiate the

effect of fiber length on the compressive behavior of the SFs (Fig. 2(e)). After the first cycle, the stress retention of SF-1 M is 83%, whereas that of SF-5 M reaches 91%, indicating excellent cyclic compressive stability. Subsequent compressive testing of SF-5 M under varying strains reveals that it maintains exceptional resilience even at a high strain of 80% (Fig. 2(f)).

The fabrication of multiscale hierarchical aerogels is achieved by introducing SiC nanowires into the interior of the SF scaffolds. To prevent damage to the SiC fibers during the CVI process, a preheating treatment (heating in air at 900 °C for 1 h) was applied to the SFs to reduce the residual carbon content within the fibers (Fig. S5 in the ESM), thereby preserving fiber structural integrity during subsequent synthesis without compromising their mechanical flexibility (Fig. S6 in the ESM). Following the incorporation of nanowires, the XRD patterns of the SFWs exhibit distinct diffraction peaks at 35.6°, 60.0°, and 71.8° (Fig. 2(a)), corresponding to the (111), (220), and (311) crystal planes of β -SiC, respectively. The emergence of these diffraction peaks can be attributed partly to the crystallization of the pretreated fibers and partly to the growth of the crystalline nanowires. The density variations of the SFWs follow a trend consistent with that of the SFs, with SFW-5 M possessing the lowest density of 0.23 g·cm⁻³. Its monolith can stand stably on a flower, as shown in Fig. 2(b). Most notably, owing to differences in siloxane sol impregnation capacity, the rate of density increase relative to the corresponding SF scaffold escalates with increasing fiber length.

The influence of fiber length on the multiscale SFWs is further

explored by analyzing the microstructural morphologies of the various samples. As shown in Fig. 3(a), apart from the primary fibers, a significant amount of flake-like SiOC forms in SFW-1 M during the CVI process. From a catalyst dispersion and mass transport perspective, dense fiber packing in short-fiber scaffolds (SFW-1 M) creates severe spatial confinement and high tortuosity. This severely restricts the deep infiltration of both the precursor and ferrocene catalyst during impregnation while also impairing gas-phase mass transfer during CVI. Consequently, nonuniform catalyst aggregation and local precursor starvation trigger isotropic deposition, resulting in unwanted SiOC flakes rather than catalyst-assisted 1D anisotropic growth. With increasing fiber length, the nanowire content in SFW-3 M increases noticeably (Fig. 3(b)). When the fiber length reaches 5 mm, SFW-5 M exhibits large voids spanning hundreds of micrometers. These open macropores effectively eliminate diffusion barriers, enabling uniform penetration of liquid precursors and unhindered mass transfer of gaseous reactants (SiO and CO) throughout the entire interior.

The formation of ultrahigh-aspect-ratio SiC nanowires within these macroscopic voids is primarily governed by a ferrocene-catalyzed vapor-liquid-solid (VLS) growth mechanism. Upon thermal decomposition, ferrocene yields uniformly dispersed Fe-based catalyst nanoparticles on the SiC fibers. Under steady precursor flux, gaseous reactants continuously dissolve into these catalyst droplets. Upon reaching supersaturation, SiC unidirectionally precipitates out along low-energy crystallographic directions. Concurrently, the unconstrained physical clearance provided by the 5 mm fiber framework prevents spatial interception, allowing continuous 1D axial propagation. Consequently, SiC nanowires with ultrahigh aspect ratios uniformly fill the interfiber gaps and envelop the SiC_f scaffold, forming a multiscale, hierarchical, all-ceramic SiC aerogel. As revealed in the magnified view of the nanowires (Fig. 3(c)), the voids created by the entangled nanowires are only approximately 1 μm , defining the microscale dimension of the multiscale architecture. At an even higher magnification, the nanowire

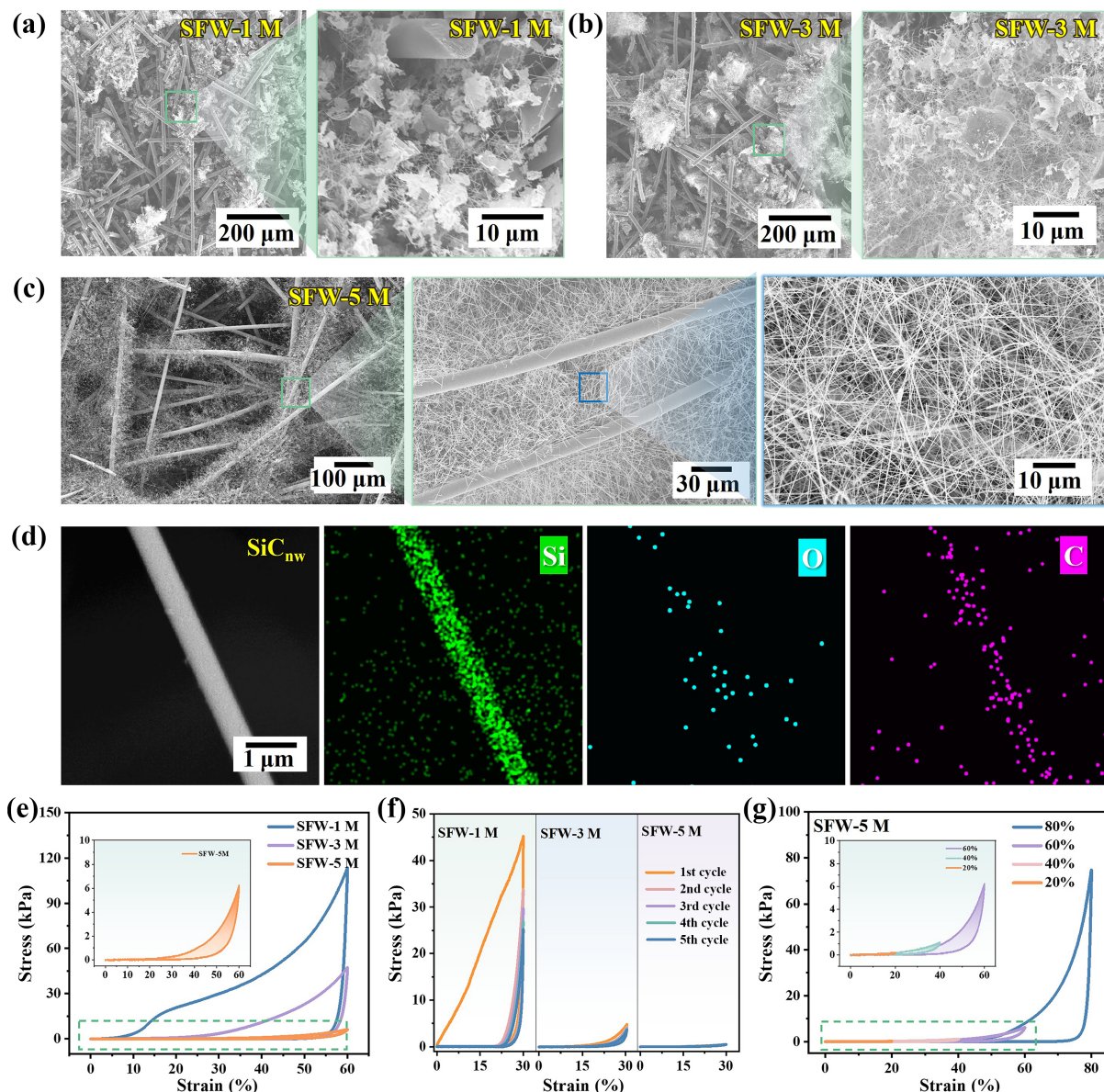


Fig. 3 Microstructural-performance correlations of the SFW aerogel series. (a–c) SEM micrographs of SFW-1 M to SFW-5 M illustrating nanowire proliferation across distinct fiber lengths. (d) High-magnification SEM micrograph and the associated EDS elemental distributions of the SiC nanowire. (e) Compressive loading–unloading stress–strain curves evaluated at 60% strain. (f) Cyclic compressive behaviors for five continuous cycles at a preset strain of 30%. (g) Compressive responses of the SFW-5 M aerogel under various maximum strains.

diameters are observed to be approximately several hundred nanometers. EDS analysis (Fig. 3(d)) confirms that the primary constituent elements are Si and C, with a negligible oxygen O content, which is highly consistent with the crystalline reflections in the XRD data. The ultrahigh aspect ratio of the SiC nanowires and the structural defects introduced during their growth collectively establish the nanoscale dimension of the multiscale architecture. This hierarchical configuration, spanning seamlessly from the millimeter to the nanometer scale, provides profound advantages for applications in thermal protection and EMW absorption.

The influence of fiber length on the mechanical properties of the SFWs exhibits a trend similar to that observed for the SFs. The compressive recovery curves at 60% strain in Fig. 3(e) indicate that the peak stress decreases with increasing fiber length. During the compression process, the SFWs likewise undergo three distinct stages Sliding, elastic deformation, and compaction. However, the presence of flake-like SiOC resulting from incomplete precursor conversion compromises the compressive resilience of the SFWs, with shorter fiber lengths yielding poorer recovery performance. Cyclic compression tests at 30% strain (Fig. 3(f)) reveal that SFW-1 M exhibits a stress retention of approximately 67% after the first cycle. In contrast, owing to the complete conversion and robust growth of the nanowires, SFW-5 M maintains exceptional resilience, achieving a stress retention of over 90% after the initial

compression. Further evaluation of SFW-5 M under varying strains demonstrates that its peak stress at 80% strain is significantly enhanced compared to that of the baseline SF-5 M scaffold (Fig. 3(g)). This mechanical superiority directly underscores the cross-scale structural synergy, where the dense nanoscale SiC nanowires act as 3D interlocking bridges and stress-transfer nodes within the macrofiber network, thereby significantly enhancing the compressive strength while greatly improving the elastic recovery rate.

3.3 EMW absorption performance

Multiscale structural design offers significant advantages in the field of EMW absorption. To demonstrate this, the EMW absorption properties of the SFs and SFWs are systematically investigated. Figures 4(a)–4(c) present the reflection loss (RL) of the bare SF scaffolds. Evidently, the RL values for samples of varying fiber lengths all remain above -10 dB, indicating poor electromagnetic wave attenuation. This is primarily attributed to the ultralow density of the SFs, which possess an air volume fraction ranging from 93% to 96%. Because air is a nonabsorbing medium, its massive presence heavily dilutes the overall mass of the scaffolds, resulting in an effective permittivity that is remarkably close to that of free space (Fig. S7 in the ESM). Specifically, for SF-5 M, the real and imaginary parts of its permittivity are nearly 1 and 0, respectively. Furthermore, the

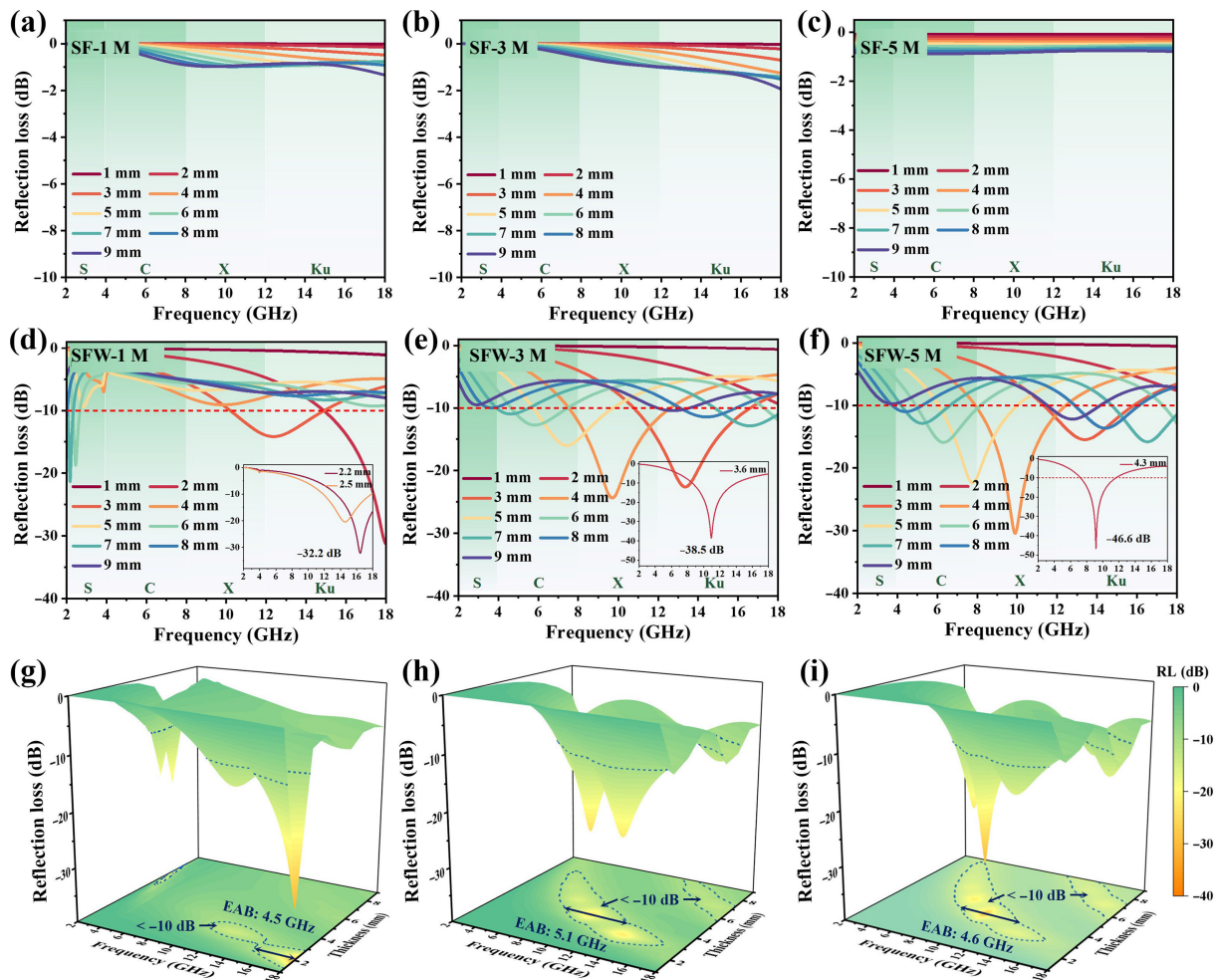


Fig. 4 EMW absorption characteristics of the SFs and SFWs within the 2.0–18.0 GHz band. (a–c) Reflection loss profiles of SF-1 M to SF-5 M at diverse thicknesses. (d–f) Reflection loss profiles of SFW-1 M to SFW-5 M at diverse thicknesses. (g–i) Three-dimensional contour plots of reflection loss for SFW-1 M to SFW-5 M as a function of frequency and thickness.

weak intrinsic electromagnetic dissipation capability of the pristine SiC materials, combined with the relatively large diameters of the fibers, results in a scarcity of heterogeneous interfaces available for internal wave reflection and scattering. Consequently, after only a few internal reflections, the electromagnetic waves easily penetrate the scaffold. Therefore, despite the exceptionally low impedance for incident waves entering the SFs, the majority of the electromagnetic energy is transmitted straight through the material due to a fundamental lack of absorbing media.

Following the integration of the smaller-scale nanowires, the EMW absorption performance of the multiscale architectures is significantly enhanced, as illustrated in Figs. 4(d)–4(f). The absorption performance of the SFWs scales with increasing fiber length. SFW-1 M exhibits an $RL_{\min}$ of -32.2 dB in the Ku-band, with an EAB of 4.5 GHz. SFW-3 M and SFW-5 M display comparable wave-attenuating capabilities. In the X- and Ku-bands, their EABs reach 5.1 and 4.6 GHz, respectively (Figs. 4(g)–4(i) and Table S3 in the ESM), with SFW-5 M achieving a lower $RL_{\min}$ of -46.6 dB. Compared to the bare SF scaffolds, the superior performance of the SFWs is driven by the introduction of the SiC nanowires. Although the air volume fraction of the SFWs remains exceptionally high ($> 90\%$), the SiC nanowires drastically expand the specific surface area of the hierarchical structure by virtue of their ultrahigh aspect ratio and a diameter only one-twentieth that of the primary SiC fibers. Consequently, the area of heterogeneous interfaces within this multiscale design is substantially increased, inducing intense interfacial polarization that drives electromagnetic attenuation without compromising impedance matching (Fig. S8 in the ESM). Because the variations in SF void geometry dictated by fiber length restrict nanowire propagation within the short-fiber scaffolds, the lower SiC nanowire loading within SFW-1 M ultimately accounts for its limited absorption capacity.

To evaluate practical engineering feasibility and validate parameter-based calculations under realistic operational conditions, a large-scale SFW-5 M panel ($180\text{ mm} \times 180\text{ mm}$) is fabricated and evaluated using the free-space arch method within an anechoic chamber. As shown in Fig. S9 in the ESM, the experimental RL profiles under normal incidence ($\theta = 90^\circ$) demonstrate remarkable congruence with the simulated predictions. Specifically, the SFW-5 M specimen delivers $RL_{\min}$ values of -30.1 and -34.8 dB under transverse electric (TE) and transverse magnetic (TM) polarizations, respectively. The corresponding EABs span 10.1 GHz (TE) and 10.4 GHz (TM), completely encompassing both the X and Ku microwave frequency bands. Additionally, the large-format component exhibits superior wide-angle wave attenuation performance, maintaining strong absorption at oblique incident angles (60° and 30°), highlighted by an $RL_{\min}$ of -49.6 dB at 60° under TM polarization. Slight deviations in peak attenuation intensity between empirical measurements and theoretical models stem predominantly from macroscopic edge diffraction phenomena and localized density variations inherent to large-scale fiber assembly. Collectively, these free-space experimental results substantiate the validity of theoretical modeling and demonstrate the exceptional electromagnetic stealth capability of the synthesized aerogel architecture.

To elucidate the underlying EMW absorption mechanisms of the multiscale SFW architectures, the complex permittivity ($\epsilon_r = \epsilon' - j\epsilon''$) of the samples is systematically evaluated. SFW-1 M exhibits anomalously high real (ϵ') and imaginary (ϵ'') parts of complex permittivity in the low-frequency range, which decline precipitously as the frequency increases, as illustrated in

Figs. 5(a) and 5(b). This phenomenon is likely attributable to its dense, restricted pore structure, which inevitably induces a severe impedance mismatch. Conversely, owing to the tailored void architectures and appropriate nanowire loadings, SFW-3 M and SFW-5 M display moderate permittivity values. As further evidenced by the dielectric loss tangent curves in Fig. 5(c), SFW-5 M achieves an optimal balance between effective electromagnetic dissipation and excellent free-space impedance matching. Meanwhile, Fig. 5(d) demonstrates that the elevated permittivity of SFW-1 M yields an exceptionally high overall attenuation constant level. Although a high attenuation constant typically implies robust electromagnetic wave attenuation capability, the excessively high permittivity triggers a severe impedance mismatch, causing incident electromagnetic waves to be primarily reflected at the sample surface rather than penetrating the interior to be absorbed. In contrast, the prominent hierarchical structure of SFW-5 M endows it with a perfectly matched attenuation constant and surface impedance (Fig. 5(e)). This synergy ensures that incident electromagnetic waves smoothly enter the aerogel. Then, the EMW energy is fully dissipated by intense multiscale polarization effects and the interconnected micronano conductive networks. The attenuation mechanisms of these hierarchical aerogels are further corroborated by Cole–Cole plots. The curve for SFW-5 M (Fig. 5(f)) exhibits several large and distinctively distorted semicircles, providing compelling evidence for the presence of multiple intense Debye relaxation processes. Conversely, the fragmented semicircles observed for SFW-1 M and SFW-3 M in Fig. S10 in the ESM confirm that although polarization occurs, the restricted nanowire content fundamentally limits the overall intensity of interfacial polarization.

The advantages of the multiscale structural design are clearly manifested in the underlying EMW absorption mechanisms. Compared to conventional single-scale SiC nanowire aerogels, which often suffer from severe surface reflection due to excessive permittivity or insufficient loss capacity from sparse networks, the hierarchical architecture drives a powerful synergy across millimeter-to-nanometer scales Fig. 5(g). Upon reaching the sample surface, incident waves smoothly penetrate the interior of the aerogel. This easy entry is enabled by excellent surface impedance matching, wherein the millimeter-scale SiC fiber scaffold maintains a low overall density and large void fraction, effectively suppressing excessive macropertmittivity.

At the macroscale, the millimeter-scale SiC fibers form a scaffold containing massive voids that span hundreds of micrometers, serving as primary propagation channels to induce continuous macroreflection and scattering. Notably, this multireflection behavior recursively occurs at a smaller scale within the dense voids of the SiC nanowires [24]. Rather than acting independently, the integrated nanowire network constructs localized 3D conductive pathways within the macroframework. Under alternating fields, these networks generate strong microcurrents to provide substantial conduction loss. From a microscopic perspective, numerous heterogeneous interfaces are established among the SiC_β , SiC_{nw} , and air phases. Under an alternating electric field, these interfaces trigger intense interfacial polarization, which serves as the primary pathway for electromagnetic energy dissipation. Concurrently, the entangled network formed by the abundant high-aspect-ratio SiC_{nw} generates microcurrents under the induced electric field, contributing to a substantial conduction loss [25]. Furthermore, at the nanoscale, numerous intrinsic defects exist within both the precursor-derived SiC fibers, and the CVI-grown SiC nanowires

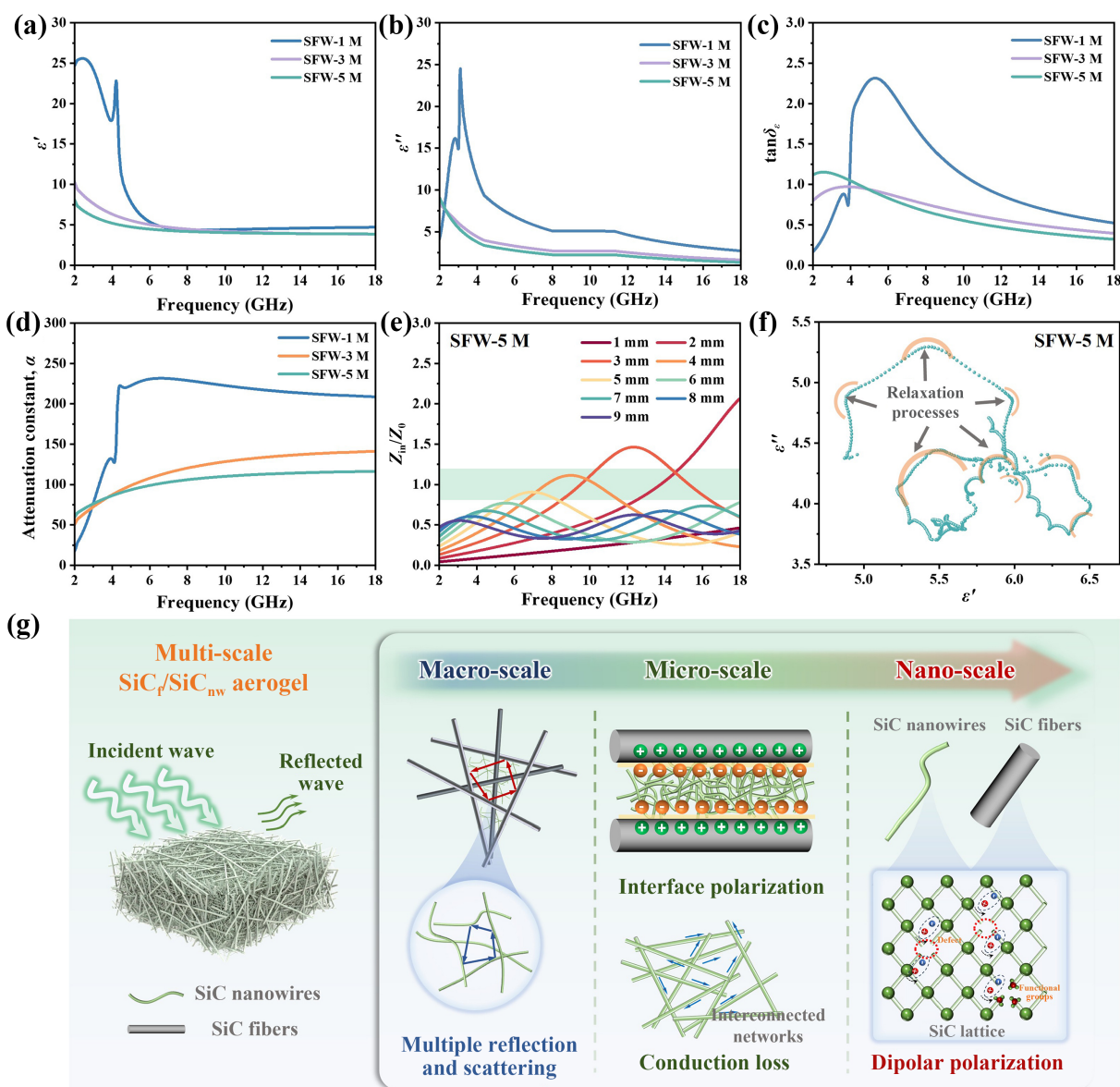


Fig. 5 Electromagnetic parameters of SFWs in the 2–18 GHz frequency range. (a, b) Real and imaginary parts of the complex permittivity. (c) Dielectric loss tangent. (d) Attenuation constant. (e) Impedance matching curves of SFW-5 M at varying thicknesses. (f) Cole-Cole plots of SFW-5 M. (g) Diagrammatic representation of the multiscale attenuation mechanisms.

generate permanent dipoles, whose reorientation further enhances dipole relaxation loss. Therefore, this multiscale architecture grants SFW aerogels a clear competitive advantage in the field of EMW absorption.

3.4 Thermal insulation performance and high-temperature stability

Thermal protection capacity serves as a vital evaluation for these advanced ceramic architectures. Accordingly, the thermal insulation properties and thermal stability of the SFs and SFWs are systematically evaluated. Thermal conductivity directly reflects the insulation capabilities of a material. Figure 6(a) illustrates the thermal conductivities of both the SFs and SFWs. Notably, SF-5 M exhibits the lowest thermal conductivity among all samples at $42 \text{ mW} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$, which is only approximately 1.5 times that of air. This outstanding performance is attributed to the ultrahigh air volume fraction within the scaffold. In contrast, shorter fibers reduce the air proportion in SF-1 M while significantly increasing the number of interfiber contact points. This promotes solid-state

thermal conduction, causing the thermal conductivity of SF-1 M to rise to $121 \text{ mW} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$. Introducing nanowires into the interfiber voids increases the overall SiC content, which naturally enhances thermal conduction. However, because a remarkably high porosity is preserved, the thermal conductivity of SFW-5 M remains at an exceptionally low level of only $64 \text{ mW} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$. To provide a more intuitive demonstration of this thermal insulation performance, a fresh flower is utilized as a visual indicator (Fig. 6(b)). When exposed directly to an alcohol lamp flame at approximately 400°C , a bare flower completely withers in less than 30 s. Conversely, protected by SFW-5 M, the flower maintains a visibly fresh state even after 3 min of continuous exposure.

To further verify the thermal protection efficacy at elevated temperatures, various 8 mm thick samples are exposed to a 1000°C butane flame. Under direct flame exposure, the back-surface temperature of the samples escalates rapidly, reaching a thermal equilibrium state after approximately 1 min. Remarkably, all samples exhibit superior thermal insulation performance. The

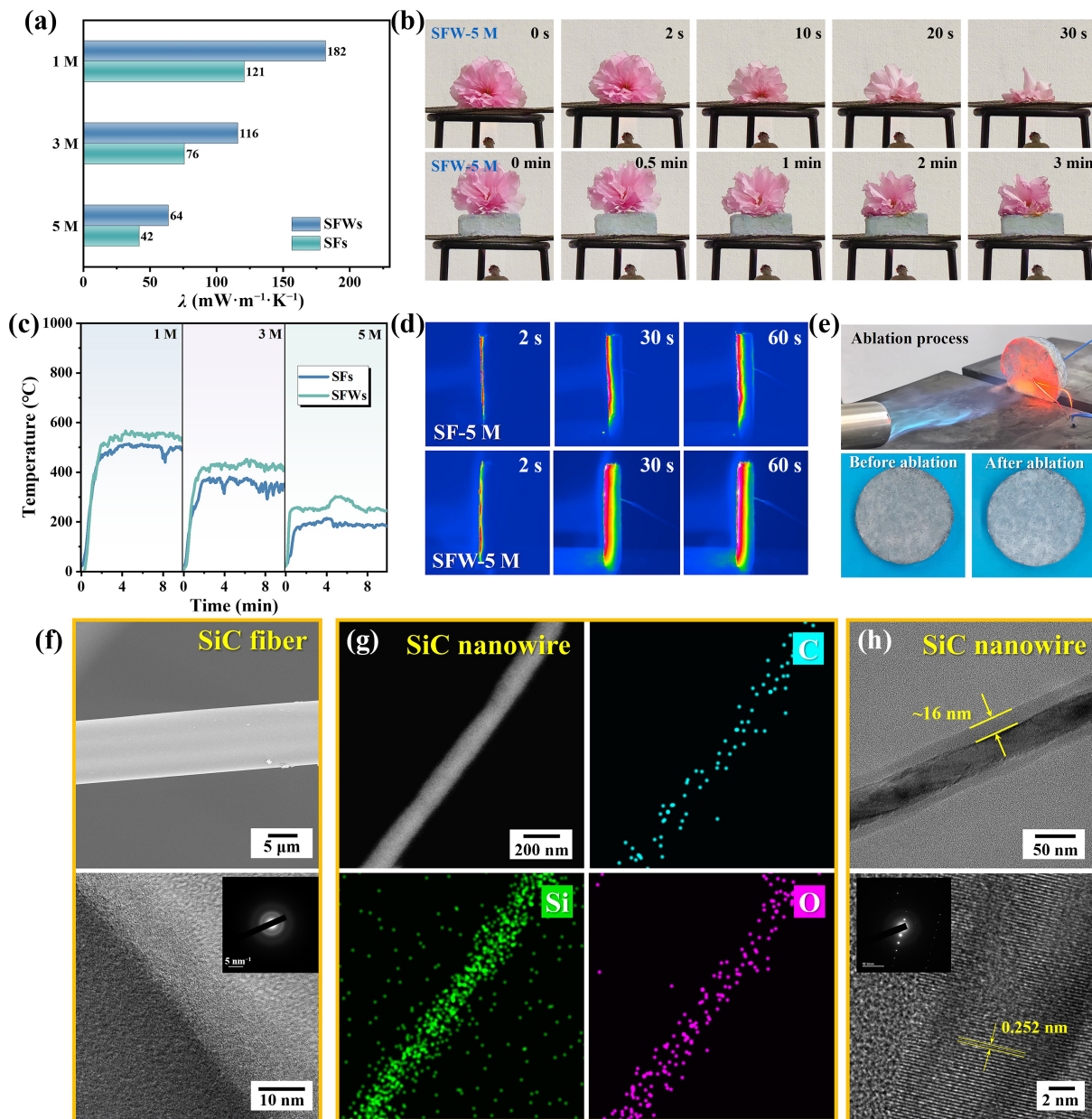


Fig. 6 Evaluation of the thermal insulation performance and high-temperature stability of the SFs and SFWs. (a) Thermal conductivity of the SFs and SFWs plotted against the fiber length. (b) Laboratory alcohol-lamp demonstration: a fresh flower shielded by an SFW-5 M block (8 mm thickness) stays unwilted for 3 min, compared to merely 30 s without shielding. (c) Back surface temperatures of the SFs and SFWs during exposure to a 1000 °C butane flame. (d) IR thermal profiles of the SFW-5 M side surfaces. (e) Ablation process under a 1000 °C flame and the surface of SFW-5 M before and after ablation. (f) SEM and TEM micrographs of SiC fibers following 1 h of air ablation at 1000 °C. (g) SEM micrographs of SiC nanowires after ablation alongside the associated EDS elemental distributions. (h) TEM micrographs of SiC nanowires after ablation.

back-surface temperature of SFW-1 M stabilizes at approximately 550 °C (Fig. 6(c)). In stark contrast, SFW-5 M, which possesses a lower thermal conductivity, maintains a back-surface temperature of only approximately 250 °C. This represents a profound 75% reduction in temperature relative to the flame source. The side-view infrared thermal imaging in Fig. 6(d) further provides direct visual confirmation of the highly localized thermal gradient established along the thickness direction of SFW-5 M.

Compared with single-scale SiC_{nw} networks, multiscale SFW architectures exhibit a prominent advantage in high-temperature thermal insulation. At lower temperatures, heat transfer is predominantly governed by gas conduction. Within this regime, the fine SiC nanowire network effectively suppresses the convective collisions of air molecules, severely restricting gas-

phase thermal conductivity. As the temperature rises, the primary mechanism of heat transfer shifts from gas conduction to thermal radiation. For single-scale nanowire scaffolds, infrared radiation with micrometer-scale wavelengths directly penetrates the material [26]. Conversely, the micrometer-sized SiC fibers within the multiscale SFWs effectively block this pathway. They induce severe scattering and absorption of infrared radiation within the same wavelength band, significantly attenuating radiative heat transfer [27]. Therefore, despite the “thermal bridging” effect introduced by the dense SiC nanowires, SFW-5 M maintains exceptional thermal insulation performance. Figure 6(e) displays the ablation testing process and the post-ablation surface state of SFW-5 M. Even after 10 min of direct flame exposure, the hierarchical structure remains completely intact. No visible

degradation, such as melting or structural collapse, is observed in the ablation area, confirming its robust flame resistance.

To evaluate the long-term thermal stability of SFW-5 M, the aerogel is exposed to a more prolonged high-temperature environment. The sample was heat-treated in a muffle furnace at 1000 °C for 1 h with a heating rate of 5 °C·min⁻¹. Figs. 6(f)–6(h) displays the post-treated microstructures of the SiC fibers and nanowires. Notably, the fiber surfaces remain remarkably smooth without any noticeable damage. TEM observation of selected fiber regions reveals the formation of a surface oxidation layer. Although XRD displays crystalline peaks after pretreatment, the low crystallinity makes the SiC crystals difficult to observe via electron diffraction or TEM (Fig. 6(f)). Nevertheless, the structural integrity remains uncompromised. Previous studies have confirmed that SF scaffolds resist structural collapse even under severe thermal exposure up to 1600 °C, highlighting their robust oxidation resistance. Figure 6(g) presents the microstructure and EDS mappings of the post-ablation SiC nanowires. The oxygen content on the nanowire surface increases sharply after the 1000 °C treatment, confirming oxide layer growth. As verified in Fig. 6(h), this amorphous oxide layer reaches a thickness of approximately 16 nm. High-resolution enlargement of the inner core region reveals well-defined lattice fringes with an interplanar spacing of approximately 0.252 nm, corresponding to the (111) crystal plane of β -SiC. This multiscale-synergistic thermal insulation performance, combined with the excellent long-term stability of the SiC scaffold, demonstrates that these multiscale SiC/SiC_{nw} architectures are highly capable of serving as reliable thermal protection systems in high-temperature and strongly oxidative environments.

3.5 Postablation EMW absorption and fire-strengthening performance

Maintaining structural reliability and functional sustainability after harsh environmental service is vital for aerogel architectures. The EMW absorption and mechanical properties of the SFs and SFWs are systematically evaluated after air ablation at 1000 °C for 1 h. Figures 7(a)–7(c) illustrate the post-ablation RL curves of the SFWs with different fiber lengths. Evidently, the EMW absorption performance of all samples is significantly enhanced after exposure. For SFW-1 M, the RL_{min} shifts from -32.2 to -51.6 dB, while its EAB expands to 7.6 GHz (Figs. 7(d)–7(f)). More importantly, the EAB of SFW-5 M widens to an impressive 11.5 GHz. This covers 72% of the entire 2–18 GHz range, almost fully spanning the C, X, and Ku bands.

An ultrathin surface oxidation layer drives this exceptionally broad EAB. The oxide layer creates abundant heterogeneous interfaces within the network. The resulting interfacial defects and structural heterogeneity act as effective polarization centers, enhancing dipole relaxation and interfacial polarization losses [28]. Furthermore, within the core-shell configuration of the formed SiC@SiO₂ structures, free charge carriers accumulate in the semiconducting SiC core, whereas the SiO₂ shell serves as a dielectric barrier. This heterogeneous structure can be conceptually modeled as a quasimicroscopic nanocapacitor network, which facilitates interfacial charge regulation and localized energy attenuation [29,30]. When these interfacial charge relaxation processes and polarization mechanisms couple with the macroscopic quarter-wavelength geometric resonance, the material exhibits two independent absorption peaks [31]. To confirm this mechanism, the theoretical quarter-wavelength curve is calculated across different thicknesses, which closely aligns with the experimental peak frequencies and confirms the geometric

resonance effect (Fig. S11 in the ESM). Because these two broad resonance valleys are closely positioned, they seamlessly merge into a continuous, uninterrupted EAB with an astonishing bandwidth of 11.5 GHz.

On the other hand, the postablation amorphous SiO₂ layer fundamentally reshapes the electrical conductivity (σ_{ac}) and impedance matching behavior. As depicted in the σ_{ac} evolution curves (Fig. S12 in the ESM), the pristine sample exhibits a relatively stable and elevated conductivity ($\sim 1\text{--}1.2\text{ S}\cdot\text{m}^{-1}$) across 2–18 GHz, which induces a noticeable skin effect and severe surface reflection. After ablation, the insulating SiO₂ shell effectively regulates the electrical conductivity of the scaffold [21]. Specifically, the formation of the oxide layer partially breaks continuous macroscopic conduction pathways, causing a marked reduction in σ_{ac} (dropping to $\sim 0.3\text{ S}\cdot\text{m}^{-1}$) at lower frequencies (2–8 GHz). This controlled attenuation of low-frequency conductivity suppresses the skin effect and dramatically optimizes surface impedance matching (Figs. 7(g)–7(i)), allowing incident electromagnetic waves to efficiently penetrate the interior. Interestingly, as the frequency increases above 9 GHz, σ_{ac} exhibits a distinct frequency-dependent dispersion behavior, rising steadily to $\sim 1.6\text{ S}\cdot\text{m}^{-1}$. This trend indicates enhanced hopping conduction of localized charge carriers across defect states and SiC@SiO₂ interfaces under high-frequency alternating fields, providing strong microwave dissipation at higher bands while preserving impedance matching at lower frequencies. As the scaffold fiber length increases, the air volume fraction within the network rises. This variation leads to a higher degree of oxidation.

Next, the postablation mechanical properties of the SFs and SFWs were investigated. Notably, all samples exhibit enhanced peak stresses during the compressive recovery process compared to their preablation states. Moreover, no catastrophic structural collapse, fracturing, or high-temperature embrittlement occurs during compression (Fig. S13 in the ESM). This resilience strongly underscores the exceptional thermal stability of these architectures. Figure 7(j) displays the smooth loading-unloading curves of the postablation SFs at 60% strain. The peak stress of SF-5 M increases from 3.3 kPa before ablation to 4.5 kPa. Shorter constituent fibers lead to a more pronounced increase in peak stress. Specifically, the peak stress of SF-1 M rises from 93.6 to 102.9 kPa. Meanwhile, the SFWs exhibit an even more remarkable upward trend (Fig. 7(k)). The postablation peak stress of SFW-5 M escalated from 6.2 to 10.5 kPa. This prominent “fire-strengthening” phenomenon is directly attributed to the thermal cross-linking of junctions between the SiC fibers and nanowires during ablation. High-magnification SEM characterization (Fig. S14 in the ESM) directly confirms this mechanism, revealing that the surface-formed amorphous SiO₂ oxidation layer softens and fuses at contact points during air ablation, generating distinct neck-like cross-linked joints at both fiber–fiber and nanowire–nanowire intersections. For the SFs, only partial cross-linking occurs at the fiber junctions due to the relatively thin SiO₂ layer. Consequently, they show only a moderate increase. Higher temperatures make this strengthening more obvious, a fact well supported by prior literature. Shorter fibers generate more contact joints, which increases cross-linking nodes and amplifies the stress. Upon incorporating the SiC nanowires, the higher network density and ultrafine diameters make junction cross-linking much easier. As a result, the SFWs yield a much stronger “fire-strengthening” effect. This performance highlights the clear benefit of the multiscale architecture. Consequently, the scalable multiscale aerogel simultaneously widens the EAB and reinforces the mechanical performance, demonstrating outstanding potential for aerospace applications in complex, harsh environments.

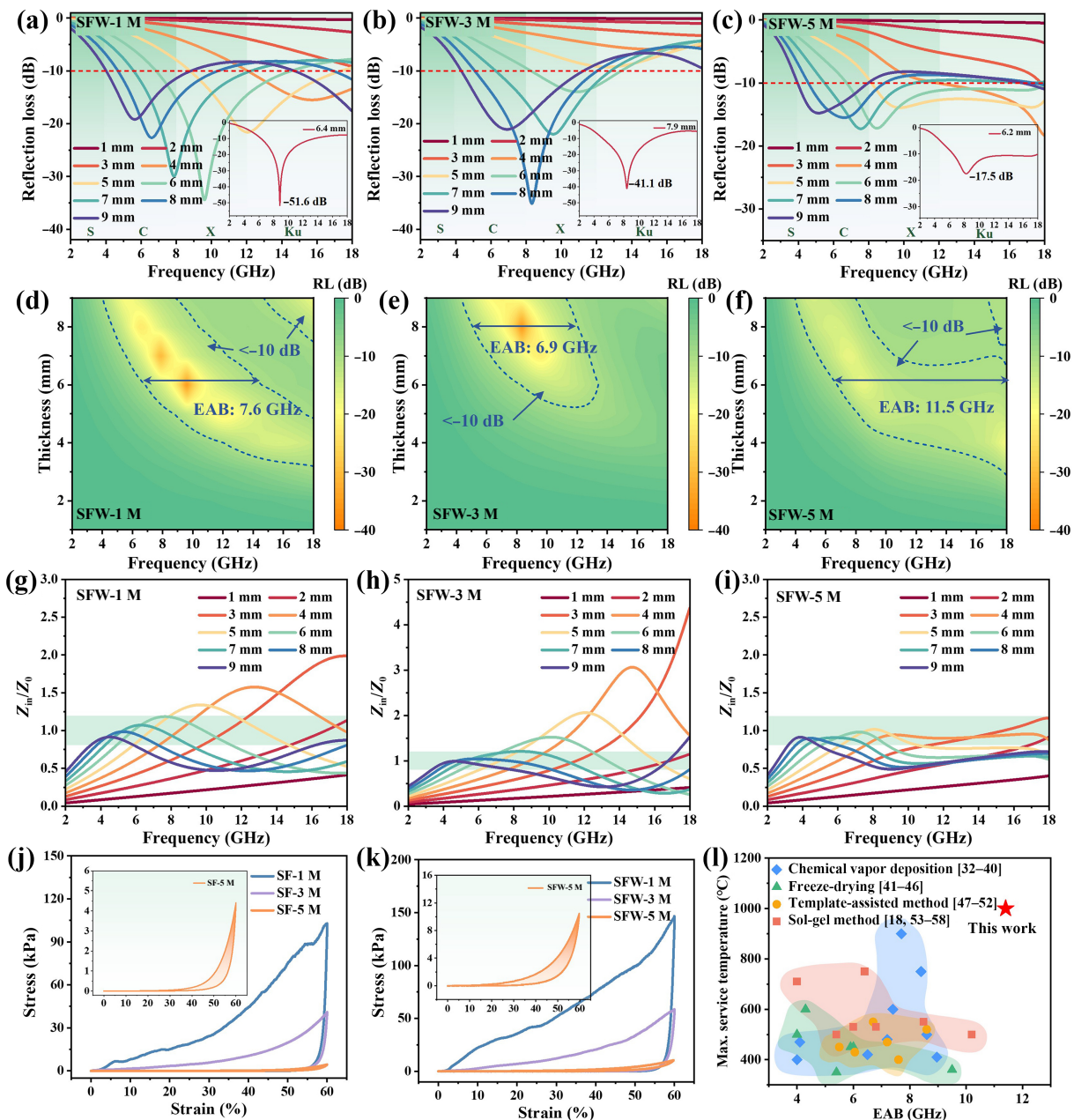


Fig. 7 Postablation EMW absorption properties and mechanical performance of the SFs and SFWs after exposure at 1000 °C in air for 1 h. (a–c) Thickness-dependent reflection loss profiles. (d–f) Thickness-dependent reflection loss 2D contour projections. (g–i) Impedance matching curves. (j–k) Stress–strain responses of SFs and SFWs during a compression–recovery loop at 60% strain. (l) Ashby chart comparing the maximum service temperature and EAB between this work and various reported materials.

As illustrated in the Ashby plot in Fig. 7(l), conventional fabrication methodologies for ceramic aerogels, including chemical vapor deposition [32–40], freeze-drying [41–46], template-assisted methods [47–52], and sol-gel [18,53–58], are heavily constrained by an inherent performance trade-off, consistently clustering in the suboptimal lower-left region with service temperatures below 900 °C and EAB restricted under 10 GHz. In sharp contrast, this work successfully breaks through these conventional material bottlenecks by occupying the ideal upper-right quadrant, simultaneously achieving an unprecedented maximum service temperature of 1000 °C and an extraordinary EAB of 11.5 GHz. To provide a more detailed and comprehensive benchmark, key performance parameters are further compared with those of state-of-the-art SiC-based, carbide-based, and ceramic aerogel absorbers in Table S4 in the ESM. This

performance breakthrough proves that multiscale design successfully overcomes the structural-functional limits of traditional single-scale materials, making it ideal for extreme aerospace environments.

4 Conclusions

In summary, we report a scalable, room-temperature ceramic papermaking strategy coupled with CVI to fabricate multiscale all-ceramic SiC_f/SiC_{nw} aerogels. This ambient pathway bridges millimeter-scale fibers and nanometer-scale nanowires, enabling the equipment-size-independent production of geometrically complex components. Modulating matrix fiber length yields an optimized scaffold (SF-5 M) that accommodates uniform proliferation of SiC nanowires. The resulting SFW-5 M aerogel

exhibits an ultralow density ($0.23 \text{ g}\cdot\text{cm}^{-3}$), low room-temperature thermal conductivity ($64 \text{ mW}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$), $RL_{\min}$ of -46.62 dB , and maintains a back-surface temperature of $\sim 250^\circ\text{C}$ under a 1000°C flame, seamlessly integrating broadband electromagnetic absorption with thermal insulation.

Crucially, high-temperature ablation (1000°C , 1 h in air) triggers a prominent “fire-strengthening” effect within the multiscale architecture. This thermal treatment simultaneously elevates the compressive peak stress and expands the effective absorption bandwidth (EAB) from 4.6 to 11.5 GHz (spanning C, X, and Ku bands). Such counterintuitive enhancement originates from thermally induced junction cross-linking and SiC@SiO₂ core-shell polarization networks. Overcoming the traditional trade-offs of single-scale aerogels, this design paradigm offers a practical route toward multifunctional thermal protection systems in extreme aerospace environments.

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Author contributions

Xiaotong Chen contributed to writing—original draft, data curation, methodology, and investigation. Jingyi Chen helped with the formal analysis and investigation. Wanxun Li contributed to data curation and review and editing. Chang Liu and Wenqing Wang contributed to data curation and investigation. Ying Li worked in supervision and writing—review & editing. Rujie He contributed to conceptualization, supervision, funding acquisition, and writing—review & editing.

Availability of data and materials

The data that support the findings of this study are available from the corresponding author upon reasonable request.

Competing interests

The authors have no competing interests to declare that are relevant to the content of this article. The author Rujie He is the Associate Editor of this journal.

Electronic Supplementary Material

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