

Graphene/SiC nanowire heterointerfaces in wood-derived carbon scaffolds for integrated electromagnetic attenuation and phase-change thermal management

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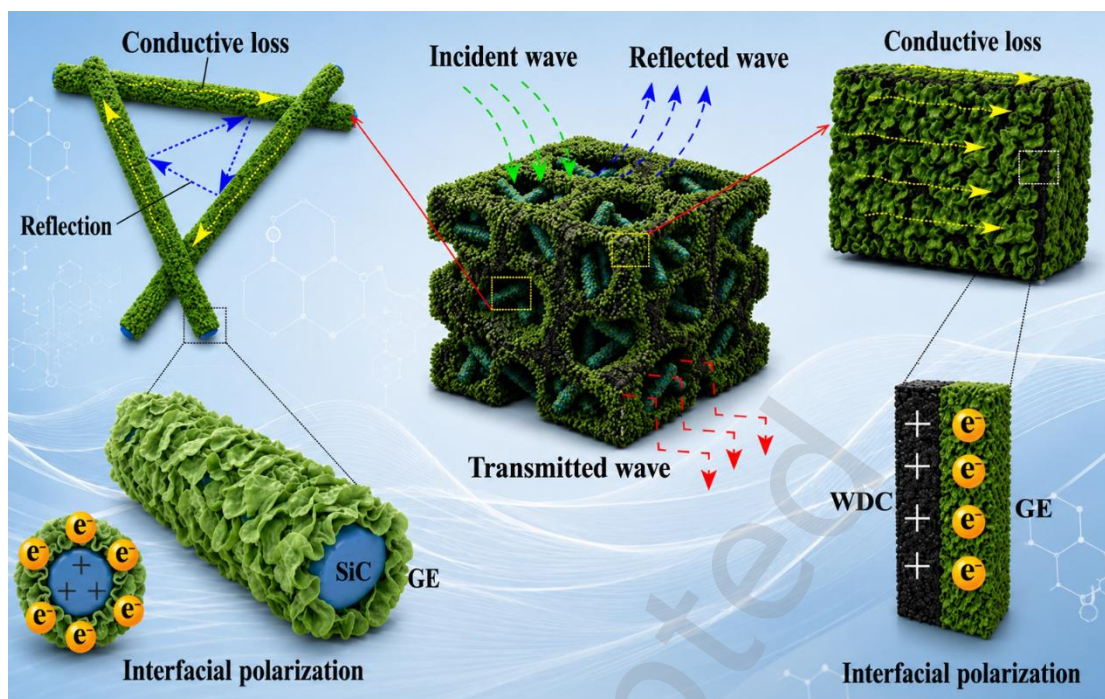


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Graphene/SiC nanowire heterointerfaces are constructed within a wood-derived carbon scaffold to enhance electromagnetic wave attenuation through interfacial polarization, dipole polarization, and conduction loss. After paraffin wax infiltration, the hierarchical scaffold enables integrated microwave absorption and phase-change thermal management with enhanced thermal transport and high energy storage efficiency.

Research Article

Graphene/SiC nanowire heterointerfaces in wood-derived carbon scaffolds for integrated electromagnetic attenuation and phase-change thermal management

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Abstract:

The progressive miniaturization and high power density of modern electronic devices have created an urgent demand for multifunctional materials that can simultaneously address electromagnetic interference (EMI) and thermal dissipation challenges. Herein, a wood-derived carbon (WDC) scaffold was sequentially modified with SiC nanowires (SiCnws) and graphene (GE) nanosheets via sol-gel/carbothermal reduction and chemical vapor deposition, followed by paraffin wax (PW) infiltration, to fabricate a multifunctional WDC-SiCnws-GE-PW phase change composite. The synergistic incorporation of SiCnws and GE constructed abundant heterointerfaces and a continuous conductive network within the hierarchically porous WDC skeleton. The WDC-SiCnws-GE absorber delivered a minimum reflection loss of -54 dB and an effective absorption bandwidth of 3.1 GHz at an ultrathin thickness of 2.1 mm. Cole-Cole analysis, together with microstructural and compositional evidence, suggested that the enhanced absorption performance is associated with a synergistic dielectric loss mechanism involving interfacial polarization at multiple heterointerfaces such as SiCnws-GE, WDC-GE, and WDC-SiCnws, dipole polarization from SiCnws stacking faults and GE edge functional groups, and conduction loss from the interconnected GE network. After PW infiltration, the resulting WDC-SiCnws-GE-PW composite exhibited outstanding thermal management performance, including a high axial thermal conductivity of 1.35 W/(m·K) (419% enhancement over WDC-PW), high melting and crystallization enthalpies (149.53 and 138.50 J/g), an energy storage efficiency of 99.35%, and rapid thermal response. This work provides a promising strategy for constructing lightweight, high-performance, multifunctional biomass-derived carbon-based

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composites for integrated EMW absorption and thermal management applications.

Keywords: electromagnetic wave absorption; wood-derived carbon; SiC nanowires; graphene; interfacial polarization; phase change thermal management

1. Introduction

The rapid advancement of modern electronic information technology, characterized by escalating integration density and operating frequencies, has led to increasingly severe electromagnetic radiation and interference issues, which threaten the operational stability of precision electronic equipment and pose potential hazards to human health [1-3]. Concurrently, the substantial heat fluxes generated in high-power components demand effective thermal management to maintain safe operating temperatures and extend device lifetime. These coexisting challenges urgently call for the development of multifunctional materials capable of simultaneously attenuating electromagnetic waves and regulating thermal loads. Consequently, developing high-performance electromagnetic wave (EMW) absorbing materials has become a research focus in the field of functional materials. Ideal EMW absorbers should simultaneously satisfy the requirements of being "thin, light, wide, and strong"—namely, thin matching thickness, low density, wide effective absorption bandwidth, and strong absorption intensity [4, 5]. Among the various candidate materials, carbon-based materials have garnered substantial attention owing to their light weight, high electrical conductivity, excellent chemical stability, and tunable structures and properties [6, 7].

In recent years, wood-derived carbon (WDC), a novel biomass-derived porous carbon material obtained by carbonizing natural wood, has attracted considerable interest [8, 9]. WDC naturally inherits the unique three-dimensional oriented porous architecture of wood. This hierarchically porous skeleton not only facilitates multiple reflections and scattering of incident EMWs within the material, thereby prolonging the transmission path, but also provides an ideal three-dimensional template for loading functional nanoscale components. For instance, Chen et al. [10] prepared Fe_3O_4 @wood-derived porous carbon composites using a delignification-impregnation-carbonization strategy, achieving a minimum reflection loss of -70.09 dB and an effective absorption bandwidth of 5.41 GHz. Dong et al. [11] constructed NiCo_2S_4 nanosheets on 3D wood-derived carbon, demonstrating the feasibility of using wood-derived carbon as a lightweight porous scaffold for loading functional nanoscale components to enhance microwave absorption. Nevertheless,

pristine WDC exhibits unsatisfactory absorption performance. The primary obstacles lie in the single dielectric loss mechanism dominated by limited conduction loss, the lack of effective polarization and relaxation processes, and the poor impedance matching that causes a substantial portion of incident EMWs to be directly reflected at the material surface [12]. Therefore, simultaneously achieving impedance matching optimization and dielectric loss enhancement while preserving the lightweight porous structural advantages of WDC remains a critical scientific challenge.

To overcome the above limitations, researchers have adopted various strategies for the functional modification of WDC. Two mainstream approaches involve introducing semiconducting nanofillers to tune the electromagnetic parameters and incorporating highly conductive nanofillers to boost dielectric loss. Silicon carbide nanowires (SiCnws), as a typical one-dimensional wide-bandgap semiconducting material, exhibit unique advantages in the field of EMW absorption owing to their high aspect ratio, abundant intrinsic defects such as stacking faults, and excellent chemical stability [13-15]. For example, Kuang et al. [16] demonstrated that the stacking fault density in SiC nanowires has a pronounced influence on dipole polarization intensity; Shi et al. [17] synthesized N-doped SiC nanowires with high-density stacking faults, showing that the synergistic effect of polarization loss induced by faults and conductivity loss caused by doping effectively enhances dielectric properties. Furthermore, the heterointerfaces formed between SiCnws and the carbon matrix can contribute interfacial polarization, thereby enriching the dielectric loss mechanisms. Meanwhile, the relatively low permittivity of SiCnws helps optimize the overall impedance matching of the composite, promoting the entry of EMWs into the material interior. Recent studies on carbide ceramic absorbers have further demonstrated that composition regulation, defect engineering, and microstructural/interface design are effective strategies for tailoring dielectric response, optimizing impedance matching, and enhancing microwave attenuation. For example, Co incorporation in high-entropy carbide ceramics can introduce additional magnetic response and dielectric-magnetic loss synergy, while high-entropy carbide-SiC biphasic ceramics fabricated via a polymer-derived ceramic route exhibit improved microwave absorption associated with optimized impedance matching, interfacial polarization, and defect-induced dipole polarization [18-20].

Graphene (GE), as a two-dimensional carbon nanomaterial, is an ideal candidate for enhancing dielectric loss owing to its ultrahigh specific surface area, exceptional

electrical conductivity, and outstanding carrier mobility [21-23]. Previous studies have demonstrated that incorporating reduced graphene oxide into porous carbon frameworks significantly elevates the conduction loss [24], and the heterointerfaces constructed between GE and semiconducting nanophases can induce strong interfacial polarization, further enriching the EMW attenuation pathways [25]. Han et al. [26] achieved hierarchical graphene/SiC nanowire networks in polymer-derived ceramics via a polymer pyrolysis process, attaining a minimum reflection loss of -69.3 dB at 10.55 GHz. The chemical vapor deposition (CVD) method enables the in-situ growth of high-quality GE within porous scaffolds, allowing uniform loading and controllable configuration of GE, and has been successfully applied to the fabrication of various carbon-based composites [27]. For example, our previous studies have demonstrated that CVD can be used to construct interconnected GE interfaces on carbon fibers and GE/nanowire hybrid networks in carbon-based composites, confirming the feasibility of forming continuous GE-containing interfacial networks in complex carbon frameworks [28,29]. However, studies that synergistically introduce SiCnws and GE into the WDC scaffold, simultaneously exploiting their complementary advantages to optimize impedance matching and enhance multiple dielectric loss mechanisms, and further elucidating the underlying synergistic attenuation mechanisms, remain scarce.

It is worth noting that in practical applications, electronic devices often face the dual challenges of electromagnetic radiation and thermal management simultaneously [30, 31]. Phase change materials (PCMs) can absorb or release substantial amounts of latent heat through solid-liquid phase transitions at nearly constant temperature, serving as effective media for passive thermal management [32]. Paraffin wax (PW), as an organic PCM, offers advantages such as high latent heat of phase change, excellent chemical stability, and low cost [33, 34]. However, the inherently low thermal conductivity of PW ($\sim 0.2 \text{ W}/(\text{m}\cdot\text{K})$) and its severe leakage problem during the molten state significantly restrict its practical application [35]. Encapsulating PW within high-thermal-conductivity porous carbon scaffolds to construct shape-stabilized composite PCMs has proven to be an effective strategy for simultaneously addressing the above issues. Song et al. [36] prepared wood-derived carbon-carbon nanotubes-paraffin wax phase change composites, achieving axial and radial thermal conductivities of 0.35 and 0.29 $\text{W}/(\text{m}\cdot\text{K})$, respectively. Zhou et al. [37] fabricated hierarchically porous wood-derived carbon scaffold/paraffin wax

composites with integrated thermal energy management and electromagnetic interference shielding capability. In addition, biomass/MXene-derived hybrid scaffolds and SiC nanowire-modified wood carbon skeletons have also been explored to improve the latent heat storage, electromagnetic shielding, anti-leakage behavior, and thermal conduction of phase-change composites [38,39]. Previous works have demonstrated that the structural design and thermal conductivity enhancement of porous scaffolds are pivotal factors for improving the overall thermal performance and multifunctionality of composite PCMs.

In light of the above considerations, this work proposes a design and fabrication strategy for multifunctional composites. First, WDC is used as the three-dimensional porous scaffold template, and SiCnws are in-situ grown within the channels via a sol-gel and carbothermal reduction process to obtain WDC-SiCnws. Subsequently, GE nanosheets are deposited onto the surface of WDC-SiCnws using a CVD method to construct a hierarchically structured WDC-SiCnws-GE composite with abundant heterointerfaces. Finally, PW is infiltrated into the above scaffold via a vacuum-assisted melt infiltration method to fabricate the WDC-SiCnws-GE-PW phase change composite that integrates efficient EMW absorption with thermal management functionality. This study systematically characterizes the microstructure, chemical composition, dielectric properties, and EMW absorption performance of the composite, conducts an in-depth analysis of the synergistic attenuation mechanisms involving interfacial polarization, dipole polarization, and conduction loss based on Debye relaxation theory, and comprehensively evaluates its thermal conductivity, thermal stability, and thermal management capability. This work aims to provide new design insights and experimental foundations for the development of lightweight, high-efficiency, and multifunctional biomass-derived carbon-based composites.

2. Experimental Section

2.1 Materials

Balsa wood was purchased from a commercial supplier and used as the precursor for wood-derived carbon (WDC). Tetraethyl orthosilicate (TEOS), anhydrous ethanol, hydrochloric acid (HCl), and deionized water were used for Si sol preparation. Methanol was used as the carbon source for graphene (GE) growth. Paraffin wax (PW, melting point ~ 55 °C) was obtained from Sinopharm Chemical Reagent Co., Ltd. Natural gas (CH_4) and argon (Ar) were of high purity. All chemicals were of

analytical grade and used without further purification.

2.2 Synthesis of WDC, WDC-SiCnws, and WDC-SiCnws-GE

Preparation of WDC. Balsa wood blocks were cut into desired dimensions and thoroughly cleaned with deionized water and ethanol. The dried blocks were carbonized in a tube furnace at 800 °C for 2 h under flowing argon with a heating rate of 5 °C min⁻¹. After natural cooling to room temperature, the resulting WDC monoliths were obtained.

Preparation of WDC-SiCnws. A Si sol was first prepared by mixing TEOS, anhydrous ethanol, HCl, and deionized water at a molar ratio of 5:10:1:5 with continuous stirring at 70 °C for 3 h, followed by cooling to room temperature. The WDC blocks were immersed in the Si sol and subjected to vacuum treatment in a vacuum oven. The vacuum was released every 30 min; after three such cycles, the samples were held at ambient pressure for 12 h, then dried at 70 °C in a forced-air oven for 24 h. The sol-impregnated WDC was subsequently placed in a vertical chemical vapor infiltration (CVI) furnace to deposit pyrolytic carbon (PyC) as the carbon source. The deposition was conducted at 1080 ± 20 °C for 20 min, with natural gas flow at 0.8–1.0 L min⁻¹ and argon flow at 2.4 L min⁻¹. After PyC deposition, the specimens were transferred to a horizontal tube furnace and heat-treated at 1600 °C for 2 h under Ar atmosphere to form WDC-SiCnws.

Preparation of WDC-SiCnws-GE. The WDC-SiCnws blocks were placed in a horizontal CVD furnace. Under a continuous vacuum of -0.09 MPa and argon atmosphere, the furnace was heated to 1100 °C. Once the temperature stabilized, the argon flow was shut off and methanol vapor (thermostated water bath at 30 °C) was introduced as the carbon source. The deposition duration of 16 h was selected based on our preliminary process screening and previous experience in constructing interconnected GE networks on nanowire-containing carbon frameworks by CVD [28, 29]. This duration was expected to provide sufficient GE coverage on the WDC-SiCnws scaffold while preserving the open porous channels required for EMW penetration and subsequent PW infiltration.

2.3 Preparation of WDC-SiCnws-GE-PW Phase Change Composites

WDC-SiCnws-GE-PW phase change composites were fabricated by a vacuum-assisted melt infiltration method. Prior to infiltration, the paraffin wax was purified by recrystallization. Briefly, 50 g of flake paraffin wax was placed in a beaker and heated at 80 °C until fully molten. After holding at 80 °C for 24 h, the melt was

cooled to room temperature. The solidified wax adjacent to the beaker wall was collected. This recrystallization cycle was repeated 3-4 times until no insoluble impurities remained in the liquid wax. The purified PW was kept at 80 °C for subsequent use.

The WDC-SiCnws-GE monoliths were immersed in the purified molten PW at 80 °C in a vacuum oven under -0.1 MPa for 2 h. When no further air bubbles escaped from the sample surface, the pressure was restored to ambient and the sample was held for an additional 2 h. This vacuum–ambient pressure cycle was repeated three times to ensure complete filling. Residual PW on the sample surface was wiped off, and the sample was cooled to room temperature to obtain WDC-SiCnws-GE-PW. For comparison, WDC-PW and WDC-SiCnws-PW were also prepared following the identical vacuum-assisted melt infiltration procedure.

2.4 Characterization

The crystalline structure was characterized by X-ray diffraction (XRD, D8 Advance, Bruker) with Cu K α radiation ($\lambda = 1.5406 \text{ \AA}$). Raman spectra were acquired on a Renishaw inVia Raman microscope with a 532 nm laser. Fourier transform infrared (FTIR) spectra were recorded on a Nicolet iS50 spectrometer using the KBr pellet method. X-ray photoelectron spectroscopy (XPS) was performed on an ESCALAB 250Xi system (Thermo Fisher) using monochromatic Al K α radiation. The morphology and microstructure were observed by field-emission scanning electron microscopy (FE-SEM, Supra55, Zeiss) and transmission electron microscopy (TEM, JEM-2100F, JEOL) equipped with energy-dispersive X-ray spectroscopy (EDS). The electromagnetic parameters (complex permittivity) were measured over the X-band (8.2-12.4 GHz) using a vector network analyzer (VNA, N5230A, Agilent) by the waveguide method. Thermal conductivity was measured with a Hot Disk TPS 2500S thermal constant analyzer. Thermogravimetric analysis (TGA) was performed on a Netzsch TG 209 F3 analyzer from 25 to 600 °C at 10 °C min⁻¹ under nitrogen. Differential scanning calorimetry (DSC) was conducted on a TA Instruments Q2500 from 0 to 80 °C at 5 °C min⁻¹. Infrared thermal images were recorded using a FLIR E8 thermal camera.

3. Results and Discussion

3.1 Structure and Morphology Characterization

Fig. 1 presents the complete synthesis process of the materials. **Fig. 2** provides

the phase composition and chemical structure characterization of the representative intermediate and final samples. As shown in **Fig. 2a**, the XRD patterns of WDC, WDC-SiCnws, WDC-SiCnws-GE, and WDC-SiCnws-GE-PW are presented together to clearly reveal the phase evolution during the stepwise fabrication process. For WDC, a broad diffraction peak appears at approximately 23° , corresponding to the (002) plane of amorphous carbon, indicating the formation of a typical non-graphitic carbon skeleton after wood carbonization. After loading SiCnws, new diffraction peaks emerge at 35.6° , 41.4° , 60.0° , and 71.8° in addition to the broad carbon peak. These peaks are indexed to the (111), (200), (220), and (311) planes of β -SiC (JCPDS No. 29-1129), confirming the successful synthesis of SiCnws within the WDC channels. Upon the introduction of GE, the XRD pattern of WDC-SiCnws-GE still retains the characteristic diffraction peaks of β -SiC, while the carbon-related diffraction signal becomes significantly intensified. This phenomenon can be mainly attributed to the deposition of CVD-derived graphitic carbon, whose (002) diffraction overlaps with the carbon peak of WDC and enhances the apparent diffraction intensity of the carbon component. After further incorporation of paraffin wax, the XRD pattern of WDC-SiCnws-GE-PW exhibits two sharp diffraction peaks at 21.5° and 23.8° , corresponding to the characteristic crystalline peaks of PW, indicating the effective filling of PW within the composite skeleton. Notably, no additional crystalline diffraction peaks are detected within the detection limit of XRD, suggesting that no obvious crystalline impurity phases are present in the as-prepared composites. However, the possible existence of amorphous components, trace secondary phases, or nanoscale heterogeneity cannot be completely excluded by XRD alone.

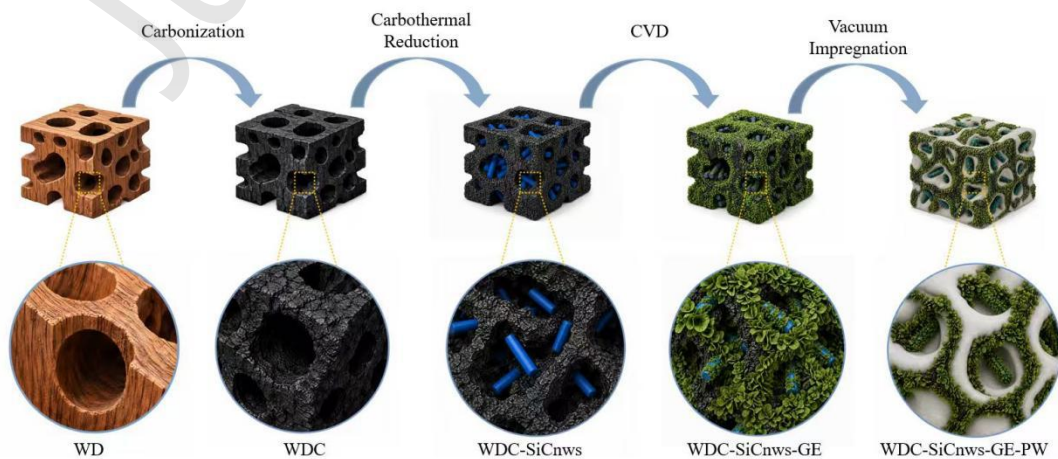


Fig. 1 Schematic illustration of the material synthesis process

Fig. 2b and **2c** present the FTIR spectra of WDC-SiCnws-GE and WDC-SiCnws-GE-PW, respectively. In the spectrum of WDC-SiCnws-GE, the absorption peak at 796 cm^{-1} corresponds to the transverse optical (TO) phonon mode of Si-C stretching vibration, while the peak at 986 cm^{-1} is attributed to the longitudinal optical (LO) phonon mode of Si-C stretching vibration. The peak at 1052 cm^{-1} is assigned to the asymmetric stretching vibration of Si-O-Si bonds, the peak at 1630 cm^{-1} arises from C=C skeletal vibration, and the broad peak at 3449 cm^{-1} originates from -OH stretching vibration. After PW infiltration (**Fig. 2c**), prominent C-H stretching vibration peaks appear at $\sim 2920\text{ cm}^{-1}$ and $\sim 2850\text{ cm}^{-1}$, accompanied by C-H bending vibration peaks at $\sim 1470\text{ cm}^{-1}$ and $\sim 720\text{ cm}^{-1}$. These are characteristic absorptions of long-chain alkanes in paraffin wax, further verifying the successful incorporation of PW into the composite.

Fig. 2d compares the Raman spectra of WDC-SiCnws and WDC-SiCnws-GE to further reveal the structural evolution of the carbon phase after CVD deposition. For WDC-SiCnws, a distinct peak at approximately 777.9 cm^{-1} is observed, which can be assigned to the transverse optical (TO) phonon mode of β -SiC, confirming the formation of SiCnws within the WDC scaffold. Meanwhile, the longitudinal optical (LO) phonon mode of β -SiC at approximately 972 cm^{-1} is not obvious, which may be related to the orientation-dependent Raman response of the SiCnws. In addition, two prominent carbon-related peaks located at approximately 1327.8 and 1568.5 cm^{-1} correspond to the D band and G band, respectively. The D band originates from defects and disordered carbon structures, while the G band is associated with the in-plane vibration of sp^2 -hybridized carbon atoms. The calculated I_D/I_G and I_{2D}/I_G values of WDC-SiCnws are 1.13 and 0.53, respectively, suggesting that the WDC/PyC-containing scaffold contains abundant defective carbon structures and relatively limited graphitic domains.

After CVD-derived GE deposition, WDC-SiCnws-GE still exhibits typical D, G, and 2D bands, while the relative intensities of these bands change noticeably. The I_{2D}/I_G and I_D/I_G values of WDC-SiCnws-GE are calculated to be 0.76 and 0.81, respectively. Compared with WDC-SiCnws, the decreased I_D/I_G value and the more pronounced 2D feature suggest that CVD deposition introduces additional graphitic nanosheet domains with graphene-like features while retaining defects and edge sites in the deposited carbon phase. These defects and edge sites can act as polarization

centers and contribute to electromagnetic wave attenuation. Combined with the veil-like sheet morphology observed by SEM/TEM, the graphitic lattice fringes, and the polycrystalline SAED pattern, the comparative Raman results support the formation of CVD-derived graphitic nanosheets within the WDC-SiCnws scaffold. For simplicity, this CVD-derived graphitic nanosheet phase is denoted as GE in the following discussion.

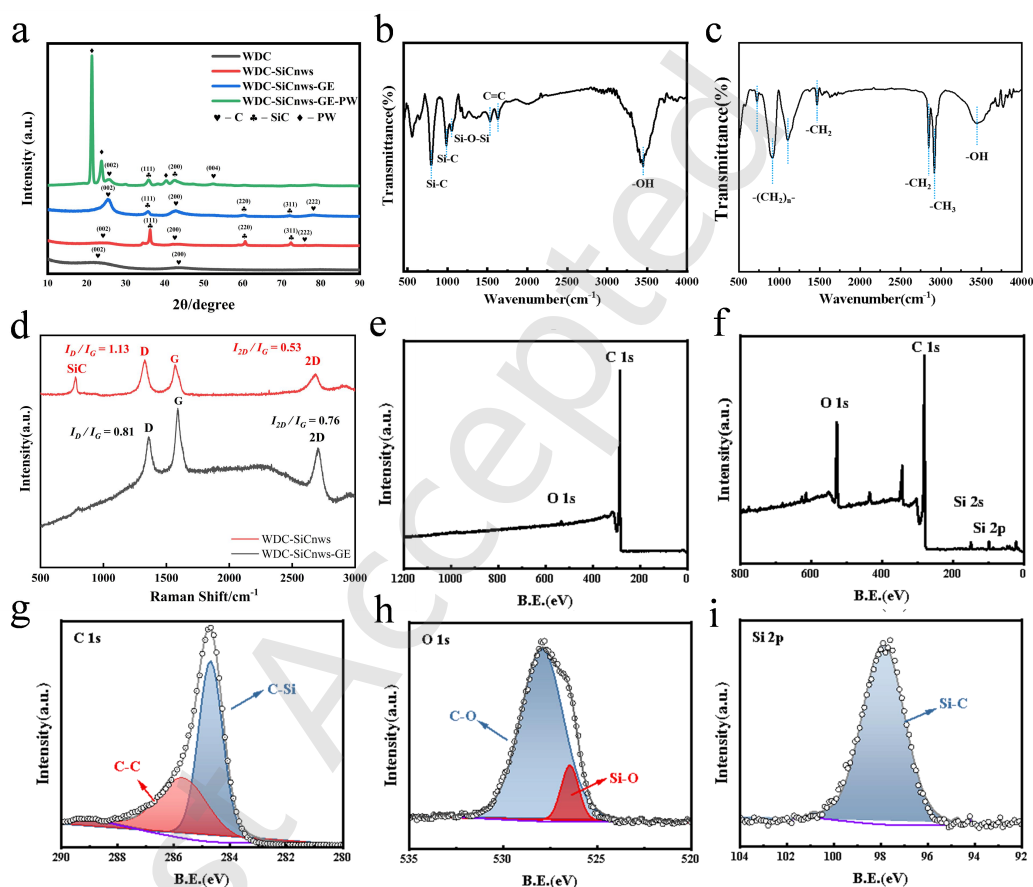


Fig. 2 (a) XRD patterns of WDC, WDC-SiCnws, WDC-SiCnws-GE, and WDC-SiCnws-GE-PW.

FTIR spectra of (b) WDC-SiCnws-GE and (c) WDC-SiCnws-GE-PW; (d) Raman spectra of WDC-SiCnws and WDC-SiCnws-GE; (e) XPS survey spectrum of WDC; XPS survey spectrum (f), C 1s spectrum (g), O 1s spectrum (h), and Si 2p spectrum (i) of WDC-SiCnws-GE.

XPS analysis was performed to further investigate the surface elemental composition and chemical states. **Fig. 2e-f** display the XPS survey spectra of WDC and WDC-SiCnws-GE, respectively. The survey spectrum of WDC primarily shows C 1s and O 1s signals, originating from the residual carbon skeleton and a small amount of oxygen-containing functional groups after wood carbonization. The survey spectrum of WDC-SiCnws-GE reveals distinct Si 2s and Si 2p characteristic peaks at 150.0 eV and 100.5 eV, respectively, confirming the presence of Si species. **Fig. 2g**

shows the high-resolution C 1s spectrum of WDC-SiCnws-GE, which can be deconvoluted into four sub-peaks at ~ 284.8 eV (C–C/C=C), ~ 286.1 eV (C–O), ~ 287.8 eV (C=O), and ~ 289.0 eV (O–C=O). The O 1s spectrum (**Fig. 2h**) can be fitted into two components at ~ 532.0 eV (C=O) and ~ 533.5 eV (C–O). The Si 2p spectrum (**Fig. 2i**) exhibits two fitted peaks at ~ 100.8 eV and ~ 101.8 eV, assigned to Si–C and Si–O bonds, respectively, further confirming the chemical state of SiCnws and suggesting partial surface oxidation.

These XRD, FTIR, Raman, and XPS results collectively demonstrate the successful stepwise construction of the WDC-SiCnws-GE-PW composite, including the carbonized WDC scaffold, crystalline SiCnws, CVD-derived GE nanosheets, and infiltrated PW.

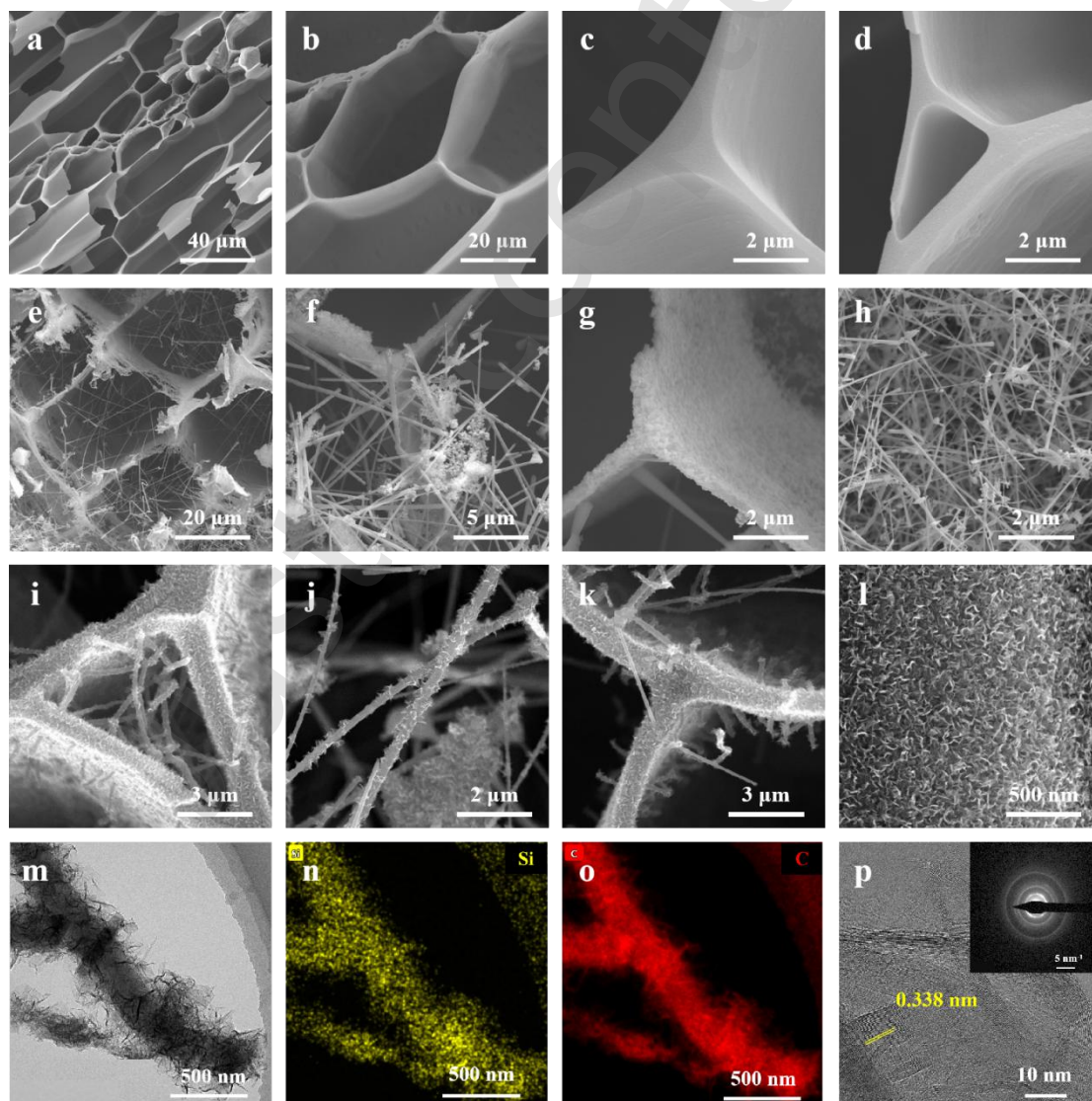


Fig. 3 SEM images of (a-d) WDC and (e-h) WDC-SiCnws; SEM images of (i) SiCnws-GE, (j) GE grown on the inner wall of SiCnws, and (k) SiCnws-GE; (l) SEM image of GE grown on the inner wall of WDC; (m) TEM image, (n,o) EDS elemental mappings, (p) HRTEM image and

corresponding SAED pattern of WDC-SiCnws-GE.

Fig. 3a-h display the SEM images revealing the microstructures of WDC and WDC-SiCnws. The pristine WDC (**Fig. 3a-d**) exhibits a three-dimensional interconnected porous structure characteristic of natural wood, with well-aligned channels and pore sizes on the order of tens of micrometers. The channel walls are relatively smooth. After modification with SiCnws (**Fig. 3e-h**), the inner walls of the channels become considerably rougher, with high-density one-dimensional nanowires uniformly covering the carbon skeleton surface and forming an intertwined network structure. These nanowires have diameters ranging from tens to hundreds of nanometers and lengths up to several micrometers.

Fig. 3i-k present the SEM images of SiCnws-GE. After CVD deposition, the graphene-like graphitic nanosheets and SiCnws interweave to form a distinctive heterostructure. Thin, veil-like GE nanosheets are observed to adhere to and conformally cover the surface of SiCnws, forming intimate interfacial contact. This conformal coating structure facilitates the construction of local electron transport pathways between SiCnws and GE and contributes to the formation of abundant heterogeneous interfaces. **Fig. 3j** further reveals the detailed morphology of GE nanosheets on the surface of SiCnws, where the thin sheets bend and conform to the curvature of the nanowire walls, suggesting close contact between the two phases. **Fig. 3l** shows that GE nanosheets are also distributed on the inner walls of WDC channels, forming a continuous and rough nanoscale interface on the originally smooth carbon skeleton.

Fig. 3m shows a representative TEM image of WDC-SiCnws-GE, in which the carbon skeleton, SiCnws, and veil-like GE nanosheets can be observed simultaneously, indicating their intimate integration. Additional TEM observation further shows that thin veil-like CVD-derived GE nanosheets are attached to the SiCnws/WDC framework, further supporting the local distribution of GE nanosheets and their intimate interfacial contact with the supporting skeleton (Fig. S1). The EDS elemental mappings (**Fig. 3n-o**) reveal a relatively homogeneous distribution of C and Si in the selected region. The C signal originates from the WDC carbon skeleton, GE nanosheets, and the carbon component in SiCnws, while the Si signal mainly derives from the SiCnws phase. In the HRTEM image (**Fig. 3p**), the interplanar spacing of approximately 0.338 nm is consistent with the (002) interlayer spacing of graphitic carbon. The SAED pattern (inset of **Fig. 3p**) exhibits polycrystalline diffraction rings.

Collectively, the above microstructural characterization results support the successful construction of a hierarchically structured WDC-SiCnws-GE composite with abundant heterointerfaces, laying a structural foundation for the subsequent electromagnetic wave absorption and thermal management performance.

3.2 Dielectric Properties and Electromagnetic Wave Absorption Performance

The electromagnetic wave absorption performance of a material is primarily governed by its dielectric properties. **Fig. 4a-c** present the frequency dependence of the real part of complex permittivity (ϵ'), imaginary part (ϵ''), and dielectric loss tangent ($\tan \delta$) for WDC, WDC-SiCnws, and WDC-SiCnws-GE over the X-band (8.2-12.4 GHz). The real permittivity ϵ' represents the capability of a material to store electromagnetic energy, while ϵ'' reflects its capacity to dissipate electromagnetic energy. As shown in **Fig. 4a**, WDC exhibits ϵ' values ranging from 8.50 to 9.02 across the X-band. After the incorporation of SiCnws, WDC-SiCnws shows a modest increase to 9.25-10.15. Notably, WDC-SiCnws-GE displays significantly enhanced ϵ' values of 11.18-12.13 throughout the measured frequency range, substantially exceeding those of both WDC and WDC-SiCnws. The pronounced enhancement in ϵ' is mainly attributed to the introduction of GE. On one hand, the highly conductive GE sheets construct a more complete conductive network within the WDC skeleton, facilitating long-range carrier migration. On the other hand, the abundant heterointerfaces formed between GE and SiCnws and between GE and WDC promote the accumulation of space charges, further strengthening the electromagnetic energy storage capability. **Fig. 4b** shows the ϵ'' curves of the three samples as a function of frequency. WDC exhibits relatively low ϵ'' values of approximately 2.60-2.82, indicating limited dielectric loss capability of the pristine carbon skeleton. WDC-SiCnws shows increased ϵ'' values of 3.13-3.75. Remarkably, WDC-SiCnws-GE achieves substantially elevated ϵ'' values of 3.88-4.33, demonstrating that the incorporation of GE significantly enhances the dielectric loss capacity. All three samples display a gradual decline in ϵ'' with increasing frequency, exhibiting typical frequency dispersion behavior, which is associated with the lag of dipole polarization response behind the alternating electromagnetic field at higher frequencies. The dielectric loss tangent, $\tan \delta$ (**Fig. 4c**), defined as the ratio of ϵ'' to ϵ' , provides a more intuitive measure of the dielectric loss efficiency. WDC shows low $\tan \delta$ values across the tested frequency range. WDC-SiCnws exhibits a moderate improvement over WDC. In contrast, WDC-SiCnws-GE maintains $\tan \delta$ values above

0.35 throughout the entire X-band, reaching a maximum of approximately 0.365 at 9.78 GHz. This trend is consistent with the variation of ϵ'' , convincingly confirming that the synergistic modification with GE and SiCnws is an effective strategy for enhancing the dielectric loss capability of WDC-based composites. **Fig. 4d** depicts the attenuation constant (α) as a function of frequency for the three samples. The attenuation constant quantifies the intrinsic attenuation capacity of a material toward incident electromagnetic waves; a larger α value indicates more pronounced energy dissipation as the wave propagates through the material. Based on transmission line theory, α can be calculated by the following equation:

$$\alpha = \frac{\pi f}{c} \sqrt{\mu' \epsilon'} \sqrt{(\tan \delta_e \delta_m - 1) + \sqrt{(1 + \tan^2 \delta_e)(1 + \tan^2 \delta_m)}} \quad (1)$$

Across the entire X-band, WDC exhibits the lowest α values, WDC-SiCnws shows a moderate increase, while WDC-SiCnws-GE delivers the highest α values among the three, indicating that the introduction of GE substantially intensifies the energy attenuation during electromagnetic wave transmission. The frequency-dependent variation of α is generally consistent with that of ϵ'' , further confirming that dielectric loss constitutes the primary mechanism for electromagnetic wave attenuation in these materials.

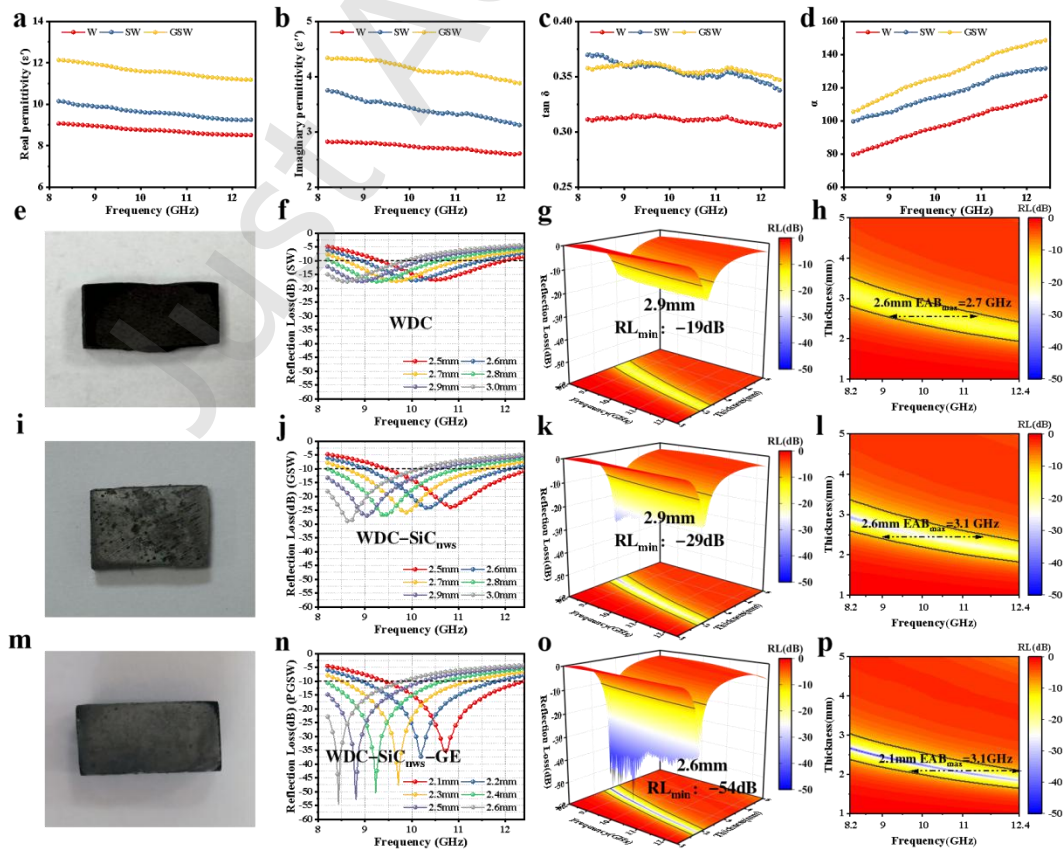


Fig. 4 (a) Real permittivity, (b) imaginary permittivity, (c) dielectric loss tangent ($\tan \delta$), and (d) attenuation constant (α) of WDC, WDC-SiCnws, and WDC-SiCnws-GE; Digital photograph, frequency-dependent RL curves, 3D RL plot, and 2D RL contour map of (e-h) WDC, (i-l) WDC-SiCnws, and (m-p) WDC-SiCnws-GE.

To visually evaluate the microwave absorption performance, the reflection loss (RL) of the three samples at varied thicknesses was calculated based on transmission line theory. **Fig. 4e-p** present the digital photographs, frequency-dependent RL curves at different thicknesses, three-dimensional RL plots, and two-dimensional RL contour maps for WDC, WDC-SiCnws, and WDC-SiCnws-GE, respectively. In general, an RL value below -10 dB corresponds to over 90% absorption of incident electromagnetic waves, and the corresponding frequency range is defined as the effective absorption bandwidth (EAB). WDC (**Fig. 4e-h**) exhibits relatively poor absorption performance, with a minimum RL of only -19 dB and a narrow EAB. Upon the introduction of SiCnws, WDC-SiCnws (**Fig. 4i-l**) demonstrates clearly improved absorption performance, achieving a minimum RL of -29 dB and an EAB of 3.1 GHz at a matching thickness of 2.6 mm. Remarkably, WDC-SiCnws-GE (**Fig. 4m-p**) delivers the most outstanding absorption performance, attaining a minimum RL of -54 dB and an EAB of 3.1 GHz at a thinner matching thickness of merely 2.1 mm, effectively covering the major portion of the X-band. Furthermore, the 2D RL contour maps reveal that the effective absorption region of WDC-SiCnws-GE can be flexibly tuned by adjusting the matching thickness (2.1-2.6 mm), demonstrating excellent designability. These results collectively indicate that the synergistic incorporation of GE and SiCnws significantly enhances the electromagnetic wave absorption capability of WDC.

To further evaluate the electromagnetic absorption performance of WDC-SiCnws-GE, representative structurally related biomass-/wood-derived absorbers were compared, as summarized in **Table S1**. These reported systems generally employ natural wood-derived porous carbon, delignified wood aerogels, ordered carbonized wood channels, or biomass-derived SiC-based porous frameworks as lightweight scaffolds, followed by the introduction of functional components such as MnO nanorods, Fe₃O₄/ZIF-67 particles, NiCo₂O₄, or SiC nanowires [8,40-42]. Compared with these systems, WDC-SiCnws-GE does not exhibit the widest EAB or the highest RL_{min} among all reported materials. However, it achieves a strong RL_{min} of -54 dB and an EAB of 3.1 GHz at a relatively thin matching thickness of 2.1 mm.

More importantly, unlike most single-function biomass-derived absorbers, the WDC-SiCnws-GE scaffold can be further infiltrated with PW to construct a phase-change thermal management composite. Therefore, the significance of this work lies in the balanced integration of efficient EMW attenuation and subsequent phase-change thermal management within one hierarchical wood-derived carbon scaffold.

To elucidate the physical mechanisms underlying the enhanced absorption performance of WDC-SiCnws-GE, **Fig. 5** further analyzes the impedance matching characteristics and Cole-Cole plots of the three samples. Impedance matching is a critical factor determining whether electromagnetic waves can effectively enter the interior of a material rather than being reflected at the surface. **Fig. 5a-f** display the normalized impedance matching characteristics of WDC, WDC-SiCnws, and WDC-SiCnws-GE, evaluated by the modulus of the normalized input impedance, $Z = |Z_{in}/Z_0|$. A Z value closer to 1 indicates easier entry of electromagnetic waves into the material. Material is considered to exhibit adequate impedance matching when the Z value falls within the range of 0.53-1.96, which is defined as the effective impedance matching region. As shown in **Fig. 5a** and **5d**, the Z values of WDC deviate considerably from unity (effective range: 0.52-1.47), indicating poor impedance matching with a substantial portion of incident waves reflected at the surface. WDC-SiCnws (**Fig. 5b** and **5e**) exhibits improved impedance matching with an effective Z range of 0.52-1.28, approaching unity to some extent. In comparison, WDC-SiCnws-GE (**Fig. 5c** and **5f**) displays Z values that are much closer to unity over a broad frequency range, with the narrowest effective range of 0.52-1.14 among the three, representing the most optimal impedance matching condition. This demonstrates that the introduction of GE does not severely deteriorate impedance matching despite its high conductivity. On the contrary, the relatively low permittivity of SiCnws effectively balances the high conductivity of GE, enabling the composite to simultaneously maintain strong dielectric loss capability and favorable impedance matching, which facilitates efficient entry and ultimate dissipation of incident electromagnetic waves. It should be noted that the present WDC-based composites feature a hierarchically porous three-dimensional skeleton, where the effective dielectric response is not a simple linear superposition of each component's contribution. Therefore, we refrain from performing quantitative deconvolution of polarization contributions based on classical Debye models developed for dense and

homogeneous bulk materials, as such an approach may introduce significant model errors in this porous system. Instead, the combination of direct microstructural evidence (heterointerfaces observed in **Fig. 3**), compositional analysis (defects and functional groups confirmed by XPS and FTIR), and the clear evolution of dielectric parameters and Cole-Cole semicircles (**Figs. 4-5**) provides robust qualitative evidence for the synergistic polarization mechanisms.

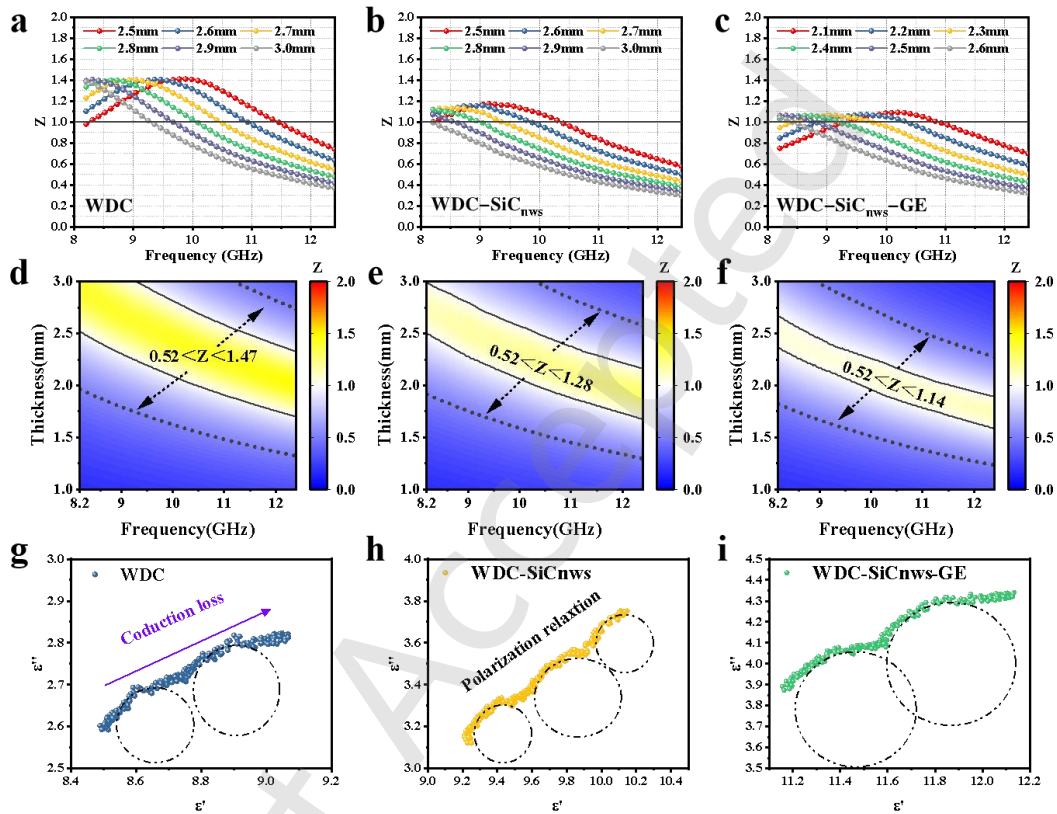


Fig. 5 Impedance matching characteristics of (a, d) WDC, (b, e) WDC-SiC_{nws}, and (c, f) WDC-SiC_{nws}-GE; Cole-Cole plots of (g) WDC, (h) WDC-SiC_{nws}, and (i) WDC-SiC_{nws}-GE.

Fig. 5g-i presents the Cole-Cole plots (ϵ'' versus ϵ') for the three samples. According to Debye relaxation theory, each Cole-Cole semicircle represents an individual polarization relaxation process. The Cole-Cole plot of WDC (**Fig. 5g**) lacks distinct semicircular features, showing only two weak and irregular small semicircles, which indicates rather weak polarization relaxation processes. These weak relaxations may originate from the residual pores, edge defects, and a small quantity of oxygen-containing functional groups on the carbon walls generated during high-temperature carbonization, which can induce weak interfacial polarization and dipole polarization under the alternating electric field. The Cole-Cole plot of WDC-SiC_{nws} (**Fig. 5h**) exhibits several more noticeable yet still irregular semicircles,

suggesting that the introduction of SiCnws contributes additional dipole polarization (from intrinsic defects such as stacking faults in SiCnws) and interfacial polarization (from WDC-SiCnws heterointerfaces). In striking contrast, the Cole-Cole plot of WDC-SiCnws-GE (**Fig. 5i**) displays multiple well-resolved semicircles, accompanied by a distinct upturned tail in the high- ϵ' region. The presence of multiple clear semicircles indicates the coexistence of multiple polarization relaxation processes in this composite: the abundant heterointerfaces formed among GE, SiCnws, and WDC (including SiCnws-GE, WDC-GE, and WDC-SiCnws interfaces) give rise to strong interfacial polarization; intrinsic defects such as stacking faults in the SiCnws lattice and residual functional groups at the edges of GE sheets serve as dipole polarization centers, generating relaxation loss under the alternating electromagnetic field. The upturned tail at the high- ϵ' end is attributed to enhanced conduction loss facilitated by the conductive network constructed by GE, the contribution of which manifests as a linear tail in the Cole-Cole representation. These results collectively support that the dielectric loss of WDC-SiCnws-GE mainly arises from the synergistic effect of interfacial polarization, dipole polarization, and conduction loss, in which interfacial polarization induced by multiple heterointerfaces may play an important role.

To further evaluate the radar stealth performance from a far-field perspective, radar cross section (RCS) simulations were performed on PEC, WDC, WDC-SiCnws, and WDC-SiCnws-GE at a matching thickness of 2.5 mm using CST Microwave Studio. **Fig. 6a1–a4** present the three-dimensional radar wave scattering patterns of the four targets at 10 GHz. The PEC plate (**Fig. 6a1**) exhibits intense specular reflection in the normal direction, with a maximum RCS value as high as 11.59 dBsm, indicative of typical total reflection behavior. The WDC target (**Fig. 6a2**) shows a reduced maximum RCS of approximately 11.47 dBsm, primarily attributed to partial attenuation and scattering of incident electromagnetic waves by its porous structure. After the introduction of SiCnws, the WDC-SiCnws target (**Fig. 6a3**) displays a maximum RCS of about 11.71 dBsm, slightly higher than that of WDC. This slight rebound may originate from the enhanced scattering effect caused by the high aspect ratio and one-dimensional morphology of SiCnws, which can increase the backscattered signal in certain directions. Notably, upon GE deposition, the WDC-SiCnws-GE target exhibits the lowest maximum RCS of 9.77 dBsm, representing a 1.82 dB reduction compared with PEC. Its scattering pattern becomes noticeably more diffuse and weaker in intensity, visually approaching the background

level. This demonstrates that the incorporation of GE not only enhances dielectric loss capability, but also effectively improves impedance matching, allowing more incident waves to enter the material and be dissipated rather than directly scattered backward.

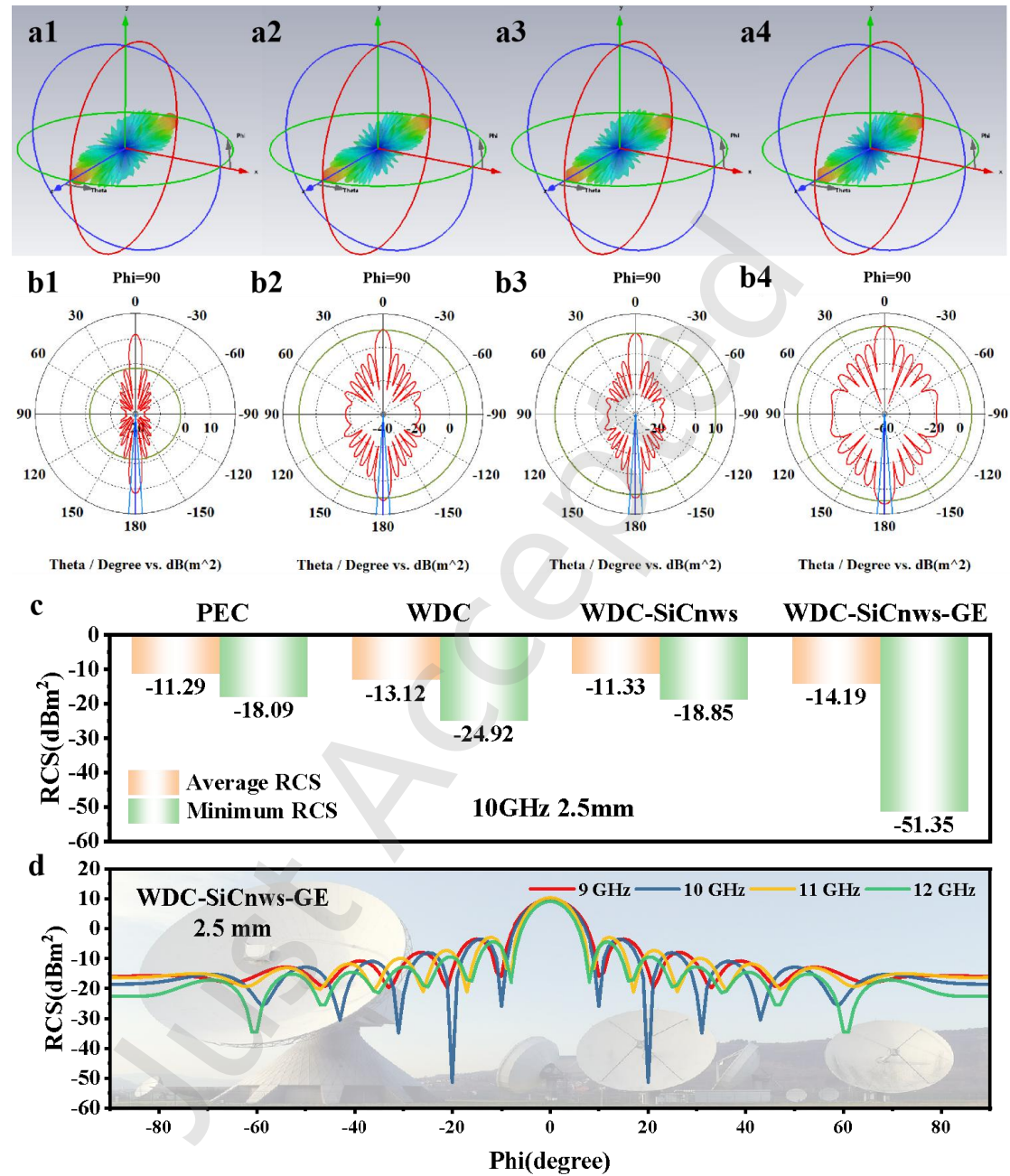


Fig. 6 (a1–a4) Three-dimensional radar scattering patterns of PEC, WDC, WDC-SiCnws, and WDC-SiCnws-GE at 2.5 mm and 10 GHz; (b1–b4) corresponding polar RCS plots; (c) comparison of average and minimum RCS values for PEC, WDC, WDC-SiCnws, and WDC-SiCnws-GE; (d) angular RCS curves of WDC-SiCnws-GE at 9, 10, 11, and 12 GHz.

Fig. 6b1–b4 show the corresponding polar RCS plots, which intuitively display the backward scattering intensity distribution of the four targets over a detection angle range of -90° to 90° . The PEC target maintains high RCS values across the entire

angular range with a relatively flat profile. The RCS values of WDC are overall lower than those of PEC. For WDC-SiCnws, the RCS is elevated at certain angles compared to WDC, consistent with the slight increase in its maximum RCS. In contrast, the WDC-SiCnws-GE target exhibits the lowest RCS values over the full angular range, with its polar curves significantly shrinking toward the center, indicating excellent radar stealth capability not only in the normal direction but also across a wide angular span.

Fig. 6c compares the average and minimum RCS values of the four targets. The average and minimum RCS of PEC are -11.29 and -18.09 dBsm, respectively. For WDC-SiCnws-GE, the average RCS decreases to -14.19 dBsm (a reduction of 2.90 dB relative to PEC), and the minimum RCS drops to -51.35 dBsm (a reduction of 33.26 dB). Both the average and minimum RCS values confirm that WDC-SiCnws-GE possesses the most superior radar scattering suppression capability among the four targets. Fig. 6d presents the angular RCS curves of WDC-SiCnws-GE at different frequencies (9, 10, 11, and 12 GHz). The RCS reduction remains stable across the entire X-band, verifying its broadband radar stealth characteristics.

In summary, the RCS simulation results strongly corroborate the excellent electromagnetic wave absorption and scattering suppression capability of WDC-SiCnws-GE from the perspective of far-field radar detection. These findings are highly consistent with the dielectric parameter analysis and reflection loss measurements discussed above, further establishing the practical application potential of this composite as a high-performance lightweight absorber.

Integrating the above analyses, **Fig. 7** presents a schematic illustration of the electromagnetic wave absorption mechanism for the WDC-SiCnws-GE composite. Incident electromagnetic waves first penetrate into the interior of the material to a large extent owing to the improved impedance matching. Once inside, the waves undergo multiple reflections and scattering within the three-dimensional porous WDC skeleton, effectively prolonging the transmission path and increasing the interaction probability with the material. At the nanoscale, the electromagnetic energy is efficiently dissipated through the following multiple mechanisms: (1) the interconnected conductive network formed by GE drives carrier migration under the alternating electromagnetic field, generating conduction loss; (2) the abundant heterointerfaces, including SiCnws-GE, WDC-GE, and WDC-SiCnws, induce space charge accumulation and redistribution due to the differences in electrical

conductivity on either side, leading to strong interfacial polarization loss; (3) intrinsic defects such as stacking faults in SiCnws and residual functional groups at GE edges act as dipole polarization centers, undergoing relaxation under the alternating electromagnetic field and contributing dipole polarization loss. The above multi-scale, multi-mechanism synergistic dissipation strategy endows WDC-SiCnws-GE with outstanding broadband electromagnetic wave absorption performance.

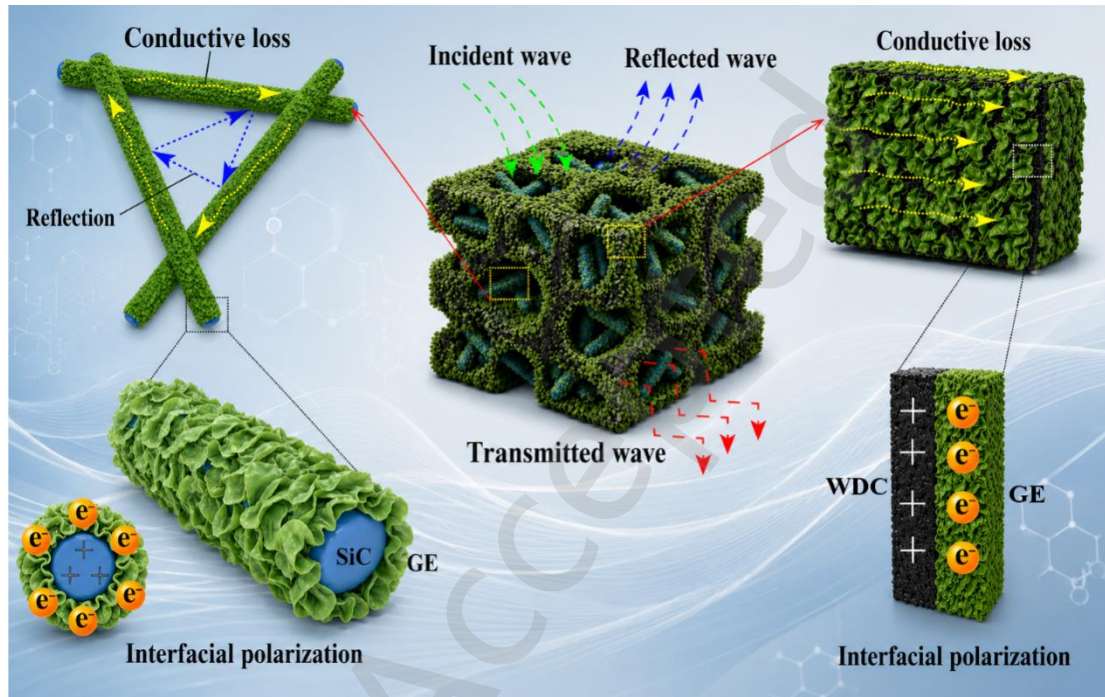


Fig. 7 Schematic illustration of the electromagnetic wave absorption mechanism for WDC-SiCnws-GE.

3.3 Thermal Management Performance

The thermal energy storage and release efficiency of phase change composites in practical applications is intimately linked to their thermal conductivity. Composites with high thermal conductivity can respond rapidly to external temperature variations, which is of critical importance for efficient thermal energy management. **Fig. 8a-d** present the SEM images of the WDC-SiCnws-GE-PW composite in the radial and axial cross-sections. It can be observed that the paraffin wax completely fills the oriented channels of the WDC-SiCnws-GE porous skeleton, exhibiting dense filling and favorable interfacial bonding without discernible phase separation or void defects. This compact filling structure is conducive to efficient heat transfer between PW and the supporting skeleton.

Fig. 8e displays the thermal conductivity of WDC-SiCnws-GE-PW measured along the axial and radial directions. The composite exhibits pronounced anisotropic

thermal transport characteristics, with an axial thermal conductivity of 1.35 W/(m·K) and a radial thermal conductivity of 1.02 W/(m·K). Compared with WDC-PW, the axial and radial thermal conductivities are enhanced by approximately 419% and 343%, respectively. Even when benchmarked against WDC-SiCnws-PW, the axial and radial thermal conductivities are still increased by approximately 61% and 26%, respectively. These comparative data indicate that the incorporation of GE plays a pivotal role in enhancing the thermal transport performance. The higher axial thermal conductivity can be mainly attributed to the naturally aligned channels inherited from the WDC scaffold, which provide continuous heat-transfer pathways along the axial direction. Meanwhile, the SiCnws and CVD-derived GE nanosheets grown on the WDC channel walls and SiCnws surfaces further bridge the carbon skeleton and reduce interfacial thermal resistance, thereby enhancing heat transport in both axial and radial directions. Therefore, the anisotropic thermal transport behavior is primarily governed by the oriented WDC channels, while the SiCnws/GE nanosheet network contributes to the overall thermal conductivity enhancement.

Thermal stability is a crucial parameter for evaluating the applicability of phase change composites in thermal energy storage and thermal regulation. **Fig. 8f-g** present the TGA and DTG curves of WDC-SiCnws-GE-PW. The TGA profile reveals a typical one-step thermal decomposition process. The onset decomposition temperature is 229.05 °C, and the maximum decomposition rate is reached at 256.87 °C. The decomposition essentially concludes at 281.49 °C, beyond which the mass stabilizes. The mass loss in this stage is entirely attributable to the thermal decomposition of PW molecular chains. The final residual mass fraction of WDC-SiCnws-GE-PW is 25.86%, from which the actual PW loading ratio within the porous skeleton is calculated to be approximately 74.14%. It is particularly noteworthy that within its intended operating temperature range (<100 °C), WDC-SiCnws-GE-PW exhibits no detectable mass loss, indicating excellent thermal stability within the operational temperature window, which fully satisfies the thermal stability requirements for practical applications.

DSC analysis was performed to characterize the phase change behavior and associated thermophysical parameters of WDC-SiCnws-GE-PW. **Fig. 8h** displays the DSC curves of WDC-SiCnws-GE-PW during the melting and crystallization processes. The phase change behavior is consistent with that of pure PW, exhibiting typical endothermic and exothermic peaks. The composite possesses a high thermal

storage density and suitable phase change temperatures, with a melting temperature (T_m) of 60.91 °C and a crystallization temperature (T_c) of 62.61 °C. The melting enthalpy (ΔH_m) and crystallization enthalpy (ΔH_c) reach 149.53 J/g and 138.50 J/g, respectively. Notably, the melting temperature of WDC-SiCnws-GE-PW is slightly lower than those of pure PW and WDC-SiCnws-PW. This may be attributed to the high thermal conductivity of the WDC-SiCnws-GE skeleton, which provides rapid heat transfer channels for PW, accelerating the phase change process and enabling heat at the composite surface to be swiftly conducted to the interior, thereby manifesting as a slight reduction in the macroscopic melting temperature.

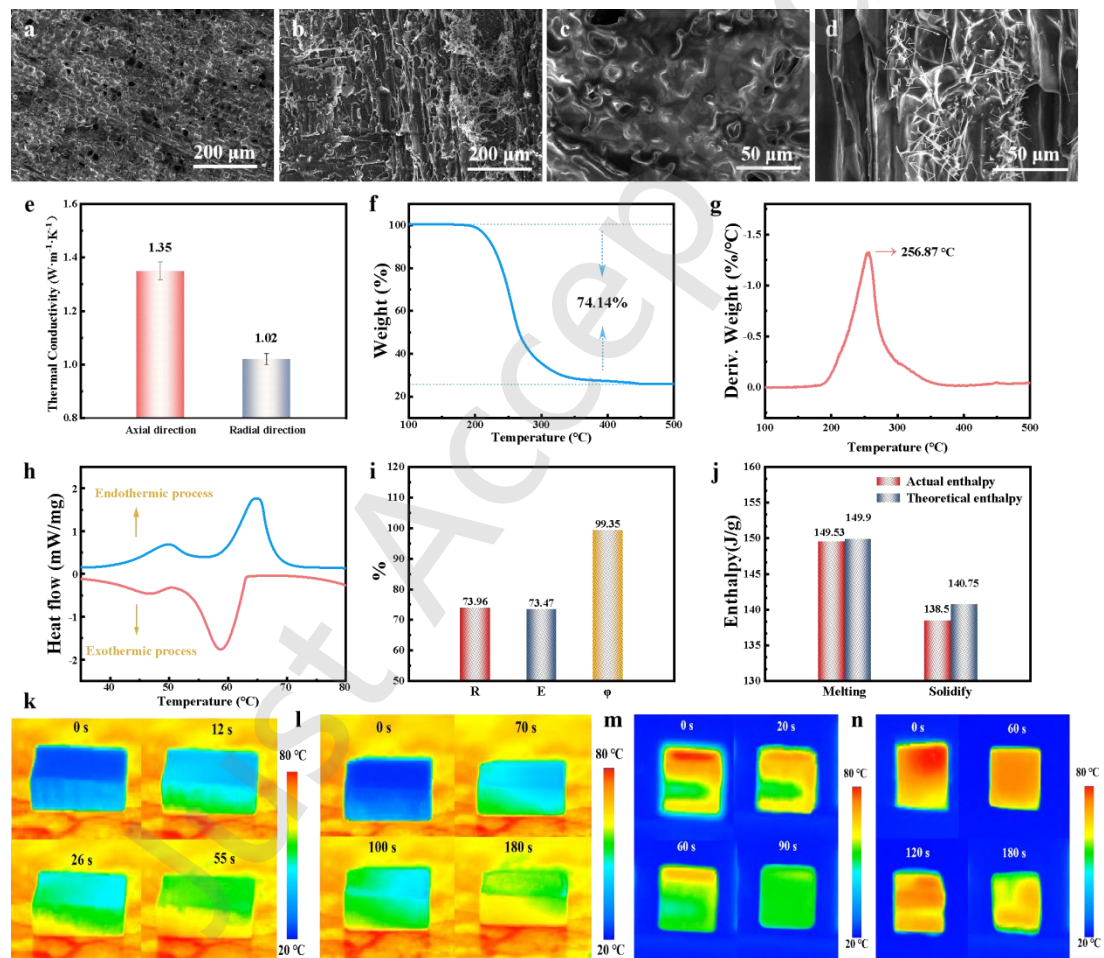


Fig. 8 SEM images of WDC-SiCnws-GE-PW: (a, c) radial section and (b, d) axial section; (e) Thermal conductivity of WDC-SiCnws-GE-PW in axial and radial directions; (f) TGA, (g) DTG and (h) DSC curves of WDC-SiCnws-GE-PW; (i) Theoretical and actual enthalpies and (j) thermal performance parameters of WDC-SiCnws-GE-PW; Infrared thermal images of (k) WDC-SiCnws-GE-PW and (l) WDC-PW during heating, and (m) WDC-SiCnws-GE-PW and (n) WDC-PW during cooling.

Fig. 8i compares the theoretical and experimentally measured enthalpies of WDC-SiCnws-GE-PW. Based on the mixture rule, the theoretical enthalpy values

during melting and crystallization are calculated to be 149.90 J/g and 140.75 J/g, respectively. The actual enthalpies are slightly lower than their corresponding theoretical values, suggesting that a fraction of PW confined within micropores does not effectively participate in the phase change process. **Fig. 8j** summarizes the key thermal performance parameters of WDC-SiCnws-GE-PW, including the impregnation ratio R (73.96%), impregnation efficiency E (73.47%), and energy storage efficiency ϕ (99.35%). The impregnation ratio R reflects the effective loading proportion of PW within the WDC-SiCnws-GE substrate, while the impregnation efficiency E describes the effective mass fraction of PW that can actually participate in thermal energy storage. The energy storage efficiency is as high as 99.35%, indicating that nearly all the loaded paraffin wax effectively contributes to the phase change thermal storage process, with only a very small fraction rendered inactive due to the confinement effect of micropores. The WDC-SiCnws-GE skeleton primarily serves dual functions in the composite: encapsulation support and thermal conductivity enhancement. It makes no direct contribution to the latent heat of phase change, as the storage and release of latent heat rely entirely on the solid–liquid phase transition of PW.

In addition to thermal conductivity enhancement, the WDC-SiCnws-GE scaffold also contributes to the shape-stabilized encapsulation of PW. The oriented porous WDC skeleton provides interconnected capillary channels for PW adsorption and physical confinement, while the in-situ grown SiCnws and CVD-derived GE nanosheets further increase the surface roughness, pore tortuosity, and interfacial restriction effect within the channels. Such a hierarchical confinement structure is beneficial for suppressing the migration of molten PW and maintaining the structural integrity of the composite during the phase-change process.

The phase-change thermal management performance of WDC-SiCnws-GE-PW was further compared with representative structurally related multifunctional phase-change composites, as summarized in **Table S2**. Reported biomass/MXene- or wood-derived carbon-based phase-change composites have demonstrated enhanced latent heat storage, EMI shielding, anti-leakage performance, or improved thermal conduction [36-39]. For example, CNC-KGM/MXene@PW shows high latent heat and excellent EMI shielding effectiveness, while wood carbon scaffold/PW composites integrate thermal energy management and EMI shielding. In addition, SiC nanowire-modified wood carbon/PW and WDC-CNTs-PW composites are

structurally close to the present work because they also employ wood-derived carbon scaffolds and secondary thermal transport networks. Compared with these systems, WDC-SiCnws-GE-PW exhibits a high axial thermal conductivity of $1.35 \text{ W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$, a radial thermal conductivity of $1.02 \text{ W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$, and high melting/crystallization enthalpies of 149.53 and $138.50 \text{ J}\cdot\text{g}^{-1}$. More importantly, the WDC-SiCnws-GE scaffold also provides strong EMW absorption before PW infiltration. These results indicate that the present composite offers a favorable balance among EMW attenuation, enhanced heat conduction, and latent heat storage.

To intuitively evaluate the thermal management capability of WDC-SiCnws-GE-PW, infrared thermal imaging was employed to record the real-time surface temperature distribution and evolution of the samples during heating and cooling. WDC-SiCnws-GE-PW and WDC-PW samples with identical thickness were simultaneously placed on a temperature-controlled hot plate, and the infrared thermal images were recorded during the heating (**Fig. 8k-l**) and cooling (**Fig. 8m-n**) processes, respectively. As shown in the infrared thermal image sequences, the surface temperature of WDC-SiCnws-GE-PW increases more rapidly than that of WDC-PW during heating, with the high-temperature region quickly spreading across the entire sample surface. Similarly, during cooling, the temperature of WDC-SiCnws-GE-PW drops more rapidly, indicating faster heat release to the surroundings. This superior thermal response rate is primarily attributable to the synergistic incorporation of GE and SiCnws, which substantially enhances the thermal conduction capability of the porous skeleton. The continuous thermal conduction network constructed by GE, together with the auxiliary thermal transport provided by SiCnws, collectively accelerates the heat transfer rate between WDC-SiCnws-GE-PW and the environment. The above results demonstrate that WDC-SiCnws-GE-PW not only possesses excellent thermal storage density and phase change stability, but also exhibits sensitive thermal responsiveness in practical application scenarios, highlighting its significant potential as an efficient thermal management material.

From the perspective of scalability and energy footprint, natural wood offers several advantages, including low cost, renewability, easy machining, and an intrinsically aligned porous architecture that can be directly converted into monolithic carbon scaffolds [8,10–12,37]. Previous studies on wood-derived carbon or biomass-derived porous scaffolds have also demonstrated their feasibility as lightweight structural platforms for electromagnetic wave absorption, thermal

management, and multifunctional phase-change composites [36–42]. In this regard, the sol infiltration, carbonization, CVD growth, and vacuum-assisted PW infiltration steps used in this work are, in principle, compatible with batch-scale processing of wood-derived monoliths with controlled dimensions.

Nevertheless, it should be acknowledged that the current fabrication route is still a proof-of-concept laboratory-scale strategy rather than a fully optimized industrial process. In particular, the high-temperature carbothermal conversion required for SiCnws formation and the relatively long CVD deposition process used for GE growth increase the overall energy consumption, processing time, and gas usage [28,29]. By contrast, the vacuum-assisted PW infiltration step is carried out at a relatively low temperature, but it still adds processing complexity when large-scale production is considered. Therefore, although the present work demonstrates a feasible strategy for constructing integrated EMW absorption and phase-change thermal management composites using a biomass-derived scaffold, further process optimization is still required to improve practical scalability. Future work will focus on reducing the synthesis temperature for SiCnws formation, shortening the CVD duration, improving carbon-source utilization, optimizing gas consumption, and developing more continuous and energy-efficient processing routes.

4. Conclusions

In summary, a multifunctional WDC-SiCnws-GE-PW composite was fabricated by sequentially growing SiCnws and GE nanosheets within a wood-derived carbon scaffold, followed by paraffin wax infiltration. The synergistic incorporation of GE and SiCnws constructed abundant heterointerfaces and a continuous conductive network, endowing the WDC-SiCnws-GE absorber with a minimum reflection loss of -54 dB and an effective absorption bandwidth of 3.1 GHz at a thickness of only 2.1 mm. The superior absorption performance was attributed to the synergistic dielectric loss mechanism combining interfacial polarization at multiple heterointerfaces, dipole polarization from SiCnws defects and GE edge groups, and conduction loss from the GE network. The CVD-derived GE nanosheets distributed on the SiC nanowires and inner walls of the WDC scaffold helped construct interconnected thermal transport pathways, thereby enabling the WDC-SiCnws-GE-PW phase change composite to achieve an axial thermal conductivity of 1.35 W/(m·K), representing a 419% enhancement over WDC-PW. Notably, the composite still maintained high melting

and crystallization enthalpies of 149.53 and 138.50 J/g, respectively, as well as an energy storage efficiency of 99.35%, demonstrating its great potential for efficient thermal buffering and passive temperature regulation under dynamic thermal environments. This work highlights the potential of hierarchically structured wood-derived carbon scaffolds with tailored heterointerfaces for addressing the coexisting electromagnetic and thermal challenges in integrated electronic systems.

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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Graphene/SiC nanowire heterointerfaces in wood-derived carbon scaffolds for integrated electromagnetic attenuation and phase-change thermal management

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Supporting Information contains:

Supporting Figure 1

Supporting Tables 1-2

Supporting References

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Supporting Figure

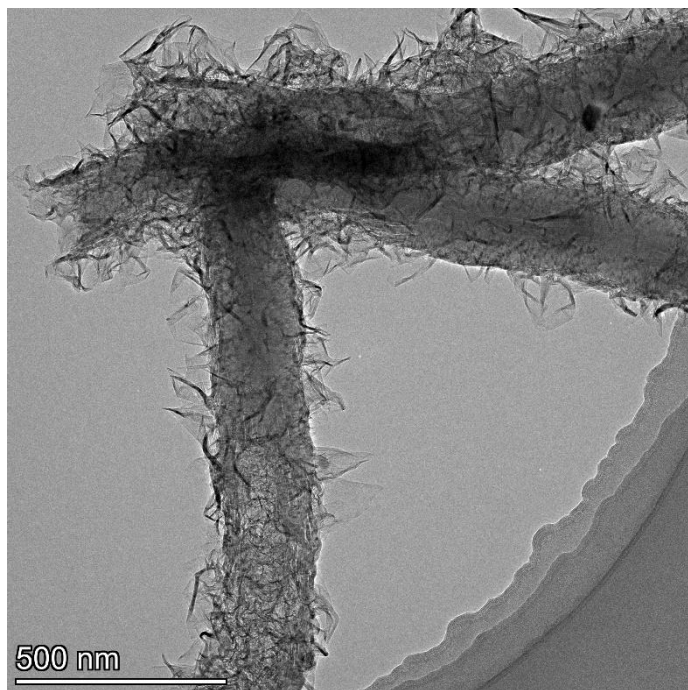


Fig. S1. Additional TEM image of the WDC-SiCnws-GE composite.

Supporting Tables

Table S1. Comparison with structurally related biomass-/wood-derived electromagnetic wave absorbers.

Materials	Biomass-/wood-derived scaffold	Functional components	RLmin / dB	EAB / GHz	Thickness / mm	Main feature	Ref.
MnOnrs/PBDC	Porous biomass-derived carbon from natural wood	MnO nanorods	−51.6	4.7	2.47	Natural wood-derived porous carbon decorated with 1D MnO nanorods; qualified bandwidth of 14.2 GHz over 1.00–5.00 mm	[1]
Fe ₃ O ₄ /ZIF-67@WA	Delignified wood aerogel	Fe ₃ O ₄ /ZIF-67 dodecahedrons	−23.4	4.5	1.5	Highly compressible, low-density wood aerogel with in-situ anchored magnetic/MOF particles	[2]
H-NiCo ₂ O ₄ @CW	Carbonized wood with ordered straight channels	NiCo ₂ O ₄	−64.0	8.08	3.62 for RLmin; 3.2 for EAB	Ordered through-channel architecture promotes multiple reflection/scattering and interface/defect polarization	[3]
3D printed porous SiCnw/SiC	Biomass-derived wood precursor processed by SLS 3D printing	SiC nanowires/SiC ceramic framework	−49.01	5.1	2.8	Biomass-derived porous ceramic skeleton with in-situ SiC nanowires; thermal conductivity of 3.21–4.99 W·m ^{−1} ·K ^{−1}	[4]
This work	Wood-derived carbon scaffold	SiCnws + CVD-derived GE nanosheets	−54	3.1	2.1	Integrates strong EMW absorption with subsequent PW-based phase-change thermal management	This work

Table S2. Comparison with structurally related multifunctional phase-change composites.

Materials	Supporting scaffold	PCM	Thermal conductivity / $\text{W} \cdot \text{m}^{-1} \cdot \text{K}^{-1}$	Phase-change enthalpy / $\text{J} \cdot \text{g}^{-1}$	Electromagnetic function	Main feature	Ref.
CNC-KGM/MXene @PW	Biomass/MXene-derived CNC-KGM/MXene hybrid scaffold	PW	Not Reported	$\Delta H_m = 215.7$	EMI SE = 45.0 dB in X-band	High latent heat, excellent EMI shielding, and solar/electro-thermal conversion capability	[5]
Wood carbon scaffold/PW	Hierarchically porous wood-derived carbon scaffold	PW	Improved compared with PW	$\Delta H_m = 122.25$; $\Delta H_c = 121.40$	Average EMI SE = 24.4 dB	Integrated thermal energy management, EMI shielding, and self-cleaning function	[6]
1300-1-SiC/C/PW	SiC nanowire-modified wood-derived carbon skeleton	PW	0.78 axial / 0.48 lateral	$\Delta H_m = 128.4$; $\Delta H_c = 128.3$	Not Reported	SiC nanowire secondary network improves anisotropic heat conduction and anti-leakage performance	[7]
WDC-CNTs-PW	Wood-derived carbon-CNT scaffold	PW	0.35 axial / 0.29 radial	$\Delta H_m = 142.02$; $\Delta H_c = 136.14$	Not Reported	CVD-grown CNT network improves thermal conduction and PW loading in WDC scaffold	[8]
This work	WDC-SiCnws-GE scaffold	PW	1.35 axial / 1.02 radial	$\Delta H_m = 149.53$; $\Delta H_c = 138.50$	EMW absorption inherited from WDC-SiCnws-GE scaffold: RL _{min} = -54 dB, EAB = 3.1 GHz	Integrates EMW attenuation, enhanced heat conduction, and latent heat storage	This work

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