

Janus-type flexible phase change composite for wearable thermal management and electromagnetic interference shielding

Hong Guo^{1,2}, Boyang Hu² (✉), Huiting Shan^{3,4}, Xiao Yang⁵, Baoan Li² (✉)

¹ Department of Energy and Chemical Engineering, Tianjin Ren'ai College, Tianjin 301636, China

² Carbon Neutrality Interdisciplinary Science Centre, Tianjin Key Laboratory of Environmental Technology for Complex Trans-media Pollution, College of Environmental Science and Engineering, Nankai University, Tianjin 300350, China

³ Department of Chemical and Biomolecular Engineering, National University of Singapore, Singapore 117585, Singapore

⁴ Joint School of National University of Singapore and Tianjin University, International Campus of Tianjin University, Fuzhou 350207, China

⁵ Hangzhou Huadian Huayuan Environment Engineering Co., Ltd., Hangzhou 310030, China

Nano Res., **Just Accepted Manuscript** • <https://doi.org/10.26599/NR.2026.94909156>

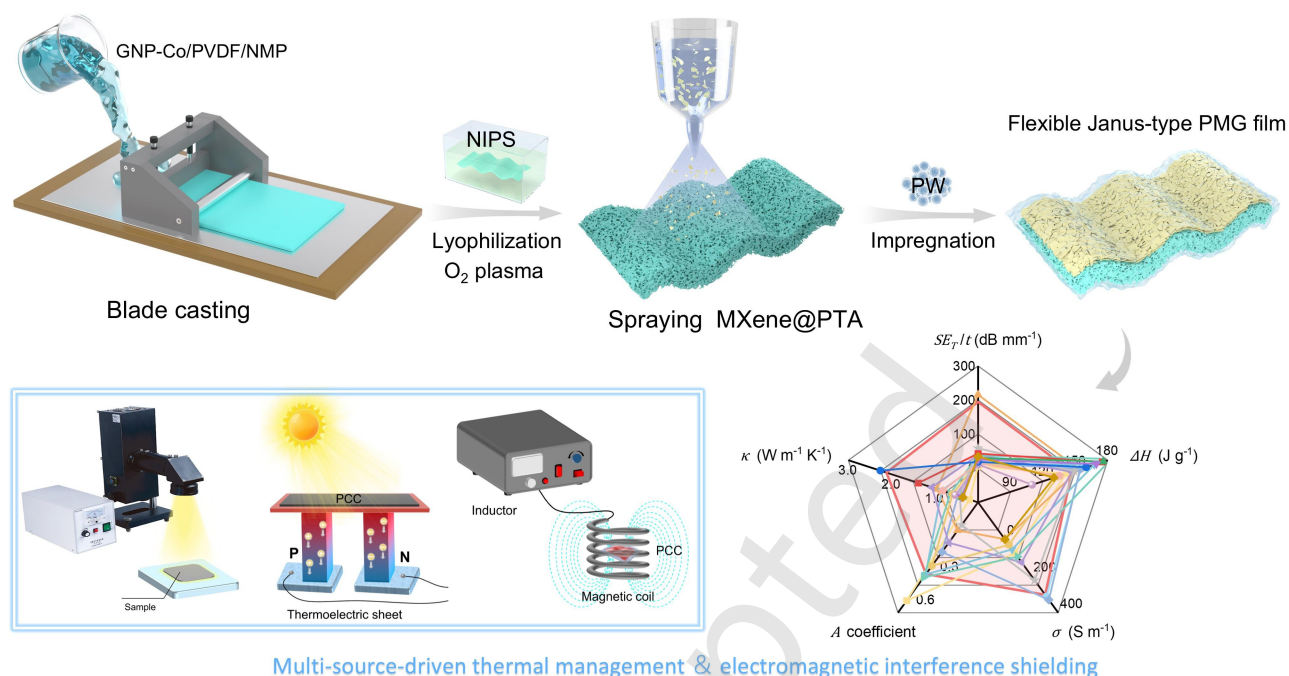
<https://www.sciopen.com/journal/1998-0124> on Aug. 31, 2026

© The Authors(s)

Just Accepted

This is a "Just Accepted" manuscript, which has been examined by the peer-review process and has been accepted for publication. A "Just Accepted" manuscript is published online shortly after its acceptance, which is prior to technical editing and formatting and author proofing. Tsinghua University Press (TUP) provides "Just Accepted" as an optional and free service which allows authors to make their results available to the research community as soon as possible after acceptance. After a manuscript has been technically edited and formatted, and the page proofs have been corrected, it will be removed from the "Just Accepted" web site and published officially with volume and article number (e.g., *Nano Research*, **2025**, 18, 94906990). Please note that technical editing may introduce minor changes to the manuscript text and/or graphics which may affect the content, and all legal disclaimers that apply to the journal pertain. In no event shall TUP be held responsible for errors or consequences arising from the use of any information contained in these "Just Accepted" manuscripts. To cite this manuscript please use its Digital Object Identifier (DOI®), which is identical for all formats of publication.

TABLE OF CONTENTS (TOC)



Low thermal conductivity, solid-state rigidity, and weak electromagnetic interference shielding effectiveness are long-standing challenges limiting PCM-based wearable thermal regulation for portable flexible electronics. In this study, scalable flexible phase change composite film is developed by a simple and low-cost strategy, which achieves multi-source-driven thermal responsiveness, EMI shielding performance, and thermal-actuated shape memory property, holding great application potential in wearable thermal management for next-generation electronic devices.

Janus-type flexible phase change composite for wearable thermal management and electromagnetic interference shielding

Hong Guo ^{1,2}, Boyang Hu ²✉, Huiting Shan ^{3,4}, Xiao Yang ⁵, Baoan Li ²✉

¹ Department of Energy and Chemical Engineering, Tianjin Ren'ai College, Tianjin 301636, China

² Carbon Neutrality Interdisciplinary Science Centre, Tianjin Key Laboratory of Environmental Technology for Complex Trans-media Pollution, College of Environmental Science and Engineering, Nankai University, Tianjin 300350, China

³ Department of Chemical and Biomolecular Engineering, National University of Singapore, Singapore 117585, Singapore

⁴ Joint School of National University of Singapore and Tianjin University, International Campus of Tianjin University, Fuzhou 350207, China

⁵ Hangzhou Huadian Huayuan Environment Engineering Co., Ltd., Hangzhou 310030, China

Received: 23 July 2026; **Revised:** 27 August 2026; **Accepted:** 31 August 2026

✉ Address correspondence to Boyang Hu, huboyang@nankai.edu.cn; Baoan Li, libaoan@nankai.edu.cn



Cite this article: Nano Research, 2026, 19, 94909156 <https://doi.org/10.26599/NR.2026.94909156>

ABSTRACT: Phase change materials (PCMs) are promising candidates for facilitating zero-energy thermal management on account of their constant phase-transition temperatures and excellent thermal storage capacity. However, low thermal conductivity, solid-state rigidity, and weak electromagnetic interference (EMI) shielding effectiveness are long-standing challenges limiting PCM-based wearable thermal regulation for portable flexible electronic devices. Herein, a Janus-type flexible phase change composite film is fabricated via a simple and low-cost phase inversion and direct spraying strategy, which incorporates an innovative combination of graphene nanoplatelets-cobalt nanoparticles@polyvinylidene fluoride (GNP-Co@PVDF) as the porous skeleton and MXene@poly tannin acid (PTA) as the sprayed layer. Resultantly, the tailored flexible composite film achieves a 960% improvement on thermal conductivity relative to paraffin, coupled with prominent EMI-shielding performance of 38.9~62.5 dB and desirable enthalpy density of 146.9 J g⁻¹. Besides, this film also exhibits leakage-resistant property, multi-source-driven thermal responsiveness, shape memory feature, and robust cyclic stability. These integrated performances substantially broaden its application potential for advanced thermal management of smart portable electronics.

KEYWORDS: flexible phase change materials (PCMs); Janus structure; wearable thermal management; electromagnetic interference (EMI) shielding; energy storage and conversion

1 Introduction

Multifunctional intelligent wearable electronic devices, with features of flexibility, multi-source powering capability, prolonged thermal energy storage and electromagnetic shielding, hold considerable application prospects in artificial skin, wearable sensors, energy conversion, and health care [1, 2]. It is of paramount importance to maintain a relatively stable and comfortable temperature for sustaining various bodily functions, especially in cold regions with insufficient sunlight [3, 4]. Traditionally, achieving a comfortable thermal environment has relied heavily on the consumption of fossil fuel-based resources for indoor heating/cooling, encompassing over 40% of the global energy expenditure [5, 6]. In response to the escalating energy crisis and environmental degradation, accelerating efficient wearable personal thermal management that performs reliably across diverse settings, whether outdoor or indoor, sunny or cloudy, is a long-sought but challenging objective [7, 8]. Concurrently, with the advances in 5G communication and integrated circuits, high-frequency portable electronics inevitably emit substantial

levels of harmful electromagnetic radiations, which diminishes operational accuracy of proximate devices and poses a threat to human health [9, 10]. Therefore, exploiting advanced multifunctional flexible films with integrated passive/active thermal management and EMI shielding capabilities is crucial for wearable electronic products.

Phase change materials (PCMs) are essential for zero-energy personal thermal management, leveraging their exceptional ability to store and release large amounts of latent heat at near-constant temperature during phase-transitions [11-13]. Through strategically choosing a PCM with an appropriate phase-transition temperature, the operational temperature of smart wearables can be effectively regulated, thereby preventing thermal runaway and mitigating performance constraints [14, 15]. Paraffin wax (PW) has garnered extensive attention due to its impressive enthalpy density, moderate phase-change temperature, cost-effectiveness, non-toxicity, remarkable thermal stability, and absence of phase separation, which render it an attractive candidate in thermal energy utilization [16-18]. Nonetheless, inherent drawbacks of liquid leakage, solid-state rigidity, and low thermal

conductivity (κ) impede its broader integration into wearable thermal management of flexible electronics [19, 20]. Furthermore, multiple clean energy sources such as solar, electrical, and magnetic fields, hold promise for thermal energy conversion and storage within PCMs, while conventional PCMs always exhibit weak thermal response to these energy inputs [21-23], greatly limiting the thermal energy utilization efficiency in advanced energy-efficient systems.

To address the challenges of liquid leakage and solid-state rigidity challenges, considerable efforts have been dedicated to exploring the shape-stabilized flexible techniques, such as elastic microencapsulation [24, 25], electrospinning [26, 27], physical/chemical conjugation [28, 29], and confinement of flexible porous substrates [30, 31]. These shape-stabilized flexible PCMs exhibit the resilience to withstand deformation and conform to the contours of uneven surfaces [32, 33], offering great potential in smart wearable applications. PCMs must not only possess flexibility but also exhibit a high κ value to enable excellent thermal regulation. The integration of high κ fillers is known to enhance the thermal charging/discharging rates of PCMs [34, 35]. While straightforward blending approach often fails to significantly augment the κ of PCMs, due to the discontinuity in thermal pathways between adjacent fillers [36, 37]. Additionally, given the ubiquity of electromagnetic radiation in modern environments, PCMs with exceptional EMI shielding performance are equally essential for safeguarding personal health [38]. To simultaneously mitigate these issues, combining PCMs with porous scaffolds for developing advanced multifunctional phase change composites (PCCs) emerges as a promising avenue [39, 40]. The versatile porous scaffold not only provides robust capillary forces to immobilize molten PCMs, but also enables advanced functionalities of PCCs, such as flexibility, multiple energy responsiveness, accelerated thermal dynamics, and superior EMI shielding performance [41, 42]. For instance, Qu et al. [43] constructed binder-free and oriented MXene- K^+ aerogel through a metal ions-induced self-assembly, followed by suction filtration and rapid freezing. Subsequent vacuum impregnation with PW yielded a PCC capable of transforming solar, electrical, and magnetic energy into thermal energy for warming, power generation, and thermal physiotherapy across multiple-scenario applications. However, the existing approaches that construct and incorporate 3D porous scaffolds are often constrained by laborious procedures, discontinuous preparation techniques, intricate fabrication processes, high energy consumption, and mold size limitations. Consequently, it is urgently needed for a simple and cost-effective strategy to produce scalable flexible phase change composite films integrating thermal management and EMI-shielding functions.

In this study, flexible Janus-type phase change composite film is tactfully customized based on a feasible phase inversion and direct spraying strategy, which successfully integrates passive/active thermal management and EMI-shielding capability. The heterogeneous GNP-Co hybrids serve as thermally conductive and microwave absorption fillers, uniformly dispersed in a porous GNP-Co@PVDF skeleton for encapsulating PW. The MXene@PTA

acts as a highly conductive reflection layer, enabling the Janus-type film to attenuate electromagnetic waves via a unique “absorption-reflection-reabsorption” mechanism. Impressively, the tailored flexible composite film achieves an excellent thermal conductivity of $2.12 \text{ W m}^{-1} \text{ K}^{-1}$, admirable EMI-shielding property of 38.9~62.5 dB, and sufficient enthalpy of 146.9 J g^{-1} , as well as leak-proof nature, thermal-driven shape recovery property, and outstanding cyclic reliability. Besides, the film demonstrates multi-source-driven thermal responsiveness, including efficient photo-thermal/photo-thermal-electric conversion for sustainable solar energy utilization, superior low-voltage Joule heating capability, and responsive magnetic-thermal conversion performance for contactless thermotherapy. In short, this work inspires a rational strategy for advancing multifunctional flexible composite PCMs, broadening their multi-scene applicability in smart wearable products.

2 Experimental section

Materials and characterizations are supplied in the Supporting Information.

2.1 Preparation of heterogeneous GNP-Co hybrids

GNPs (1 g) were initially subjected to acid treatment in a mixture of nitric acid (40 mL) and sulfuric acid (120 mL) at 60°C for a duration of 5 h. Subsequently, 0.4 g of the acid-treated GNPs, 0.2 g of polyvinylpyrrolidone (PVP), and 0.582 g of $\text{Co}(\text{NO}_3)_2 \cdot 6\text{H}_2\text{O}$ were added into 100 mL of methanol. Then mixture was homogenized using an ultra-turrax. Meanwhile, 0.675 g of 2-methylimidazole (2-MeIm) was ultrasonically dispersed in 100 mL of methanol. The two methanolic dispersions were then combined and stirred continuously for 24 h. After repetitively centrifuging and washing with methanol, the acquired precipitate was dried at 60°C and denoted as GNP@ZIF-67. Ultimately, the GNP-Co was synthesized through the pyrolysis of GNP@ZIF-67 at 800°C under an argon atmosphere for 2 h.

2.2 Preparation of porous GNP-Co@PVDF skeleton

The GNP-Co@PVDF skeleton was fabricated using a non-solvent induced phase separation (NIPS) method. Initially, viscous PVDF solution (10 wt%) was prepared by dissolving PVDF into N-methyl pyrrolidone (NMP) at 80°C for 2 h. A precise quantity of GNP-Co was then dispersed in PVDF/NMP and stirred uniformly at 60°C for 6 h, followed by degassing with a planetary gravity mixer. A series of GNP-Co@PVDF was prepared with varying mass ratios, namely 0/1, 0.25/1, 0.5/1, 0.75/1, and 1/1. Afterward, above mixture was scraped on the glass substrate to prepare wet films using a doctor-blade coating apparatus, followed by immersion into a water bath for phase separation. Finally, the wet films were freeze-dried (-80°C , $< 1 \text{ Pa}$). The resultant porous GNP-Co@PVDF skeleton was denoted as G_x , where x is the weight ratio of GNP-Co and PVDF.

2.3 Fabrication of Janus-type phase change composite film

Few-layer MXene was synthesized through a selective chemical etching strategy, as previously reported [44]. Then, 0.2 g of MXene was dispersed in 200 mL of Tris-hydrochloric buffer ($\text{pH}=8.5$, 10 mM), followed by adding 0.2 g of tannic acid (TA) and continuously stirring the mixture for 24 h at

room temperature. Superfluous TA was removed by repetitively centrifuging and washing with deionized water. The acquired precipitate was then lyophilized and denoted as MXene@PTA. To facilitate the formation of a hydrogen-bonded interface between the GNP-Co@PVDF skeleton and MXene@PTA, the GNP-Co@PVDF skeleton was treated by oxygen plasma at 100 W for 1 min. Subsequently, an aqueous dispersion of MXene@PTA (5 mg mL^{-1}) was sprayed onto the plasma-treated GNP-Co@PVDF substrate with an airbrush to acquire a hydrogen-bond-connected MXene@PTA/GNP-Co@PVDF skeleton, designated as MG_x. Finally, the MG_x skeleton was impregnated into melted PW molecules at 90 °C, and vacuum-degassed for 8 h to ensure complete impregnation. After removing the excess PW on the sample surface, the flexible PW/MG_x (PMG_x) film was obtained.

3 Results and discussion

3.1 Design strategy and microstructure characterization

Herein, we introduce a feasible phase inversion and direct spraying strategy to fabricate a scalable Janus-type flexible PMG_x film, which encompasses preparing porous GNP-Co@PVDF skeleton, utilizing MXene@PTA via complementary hydrogen-bonding interactions, and the subsequent impregnation with PW (Fig. 1a, b). Briefly, the heterogeneous GNP-Co hybrids are first synthesized by in

situ anchoring ZIF-67 onto acidified GNPs through electrostatic adsorption, subsequently treated with high-temperature pyrolysis (Fig. S1). The resultant GNP-Co, serving as thermally conductive/microwave absorption fillers, is integrated with PVDF to form porous GNP-Co@PVDF skeleton with a novel NIPS strategy. The G_x surface is then hydroxylated by oxygen plasma pretreatment and sprayed coating with MXene@PTA, subsequently impregnating PW to fabricate Janus-type PMG_x film. The commercially available GNPs exhibit a typical 2D sheet-like morphology with 3 nm in thickness and 2 μm in diameter. After acid treatment, the GNPs are functionalized with carboxylic groups (–COOH), facilitating the adsorption of metal Co²⁺ ions on their surface through electrostatic interactions. The self-assembly between these metal ions and 2-MeIm ligands effectively anchors a multitude of ZIF-67 units onto the GNP surface. As shown in Fig. 2b, Co nanoparticles with size of ~50 nm are uniformly dispersed and anchored on the GNPs, undergoing a high-temperature pyrolysis. The XRD pattern of GNP@ZIF-67 reserves all characteristic peaks of GNPs and ZIF-67, without any peak position shifts (Fig. 2a), substantiating the successful anchoring of high-purity ZIF-67 on the GNP surface. The diffraction peaks of GNP-Co at $2\theta = 44.3$ and 26.5° correspond to the (111) lattice plane of Co nanoparticles and (002) lattice plane of graphite carbon, indicating the successful synthesis of GNP-Co hybrids.

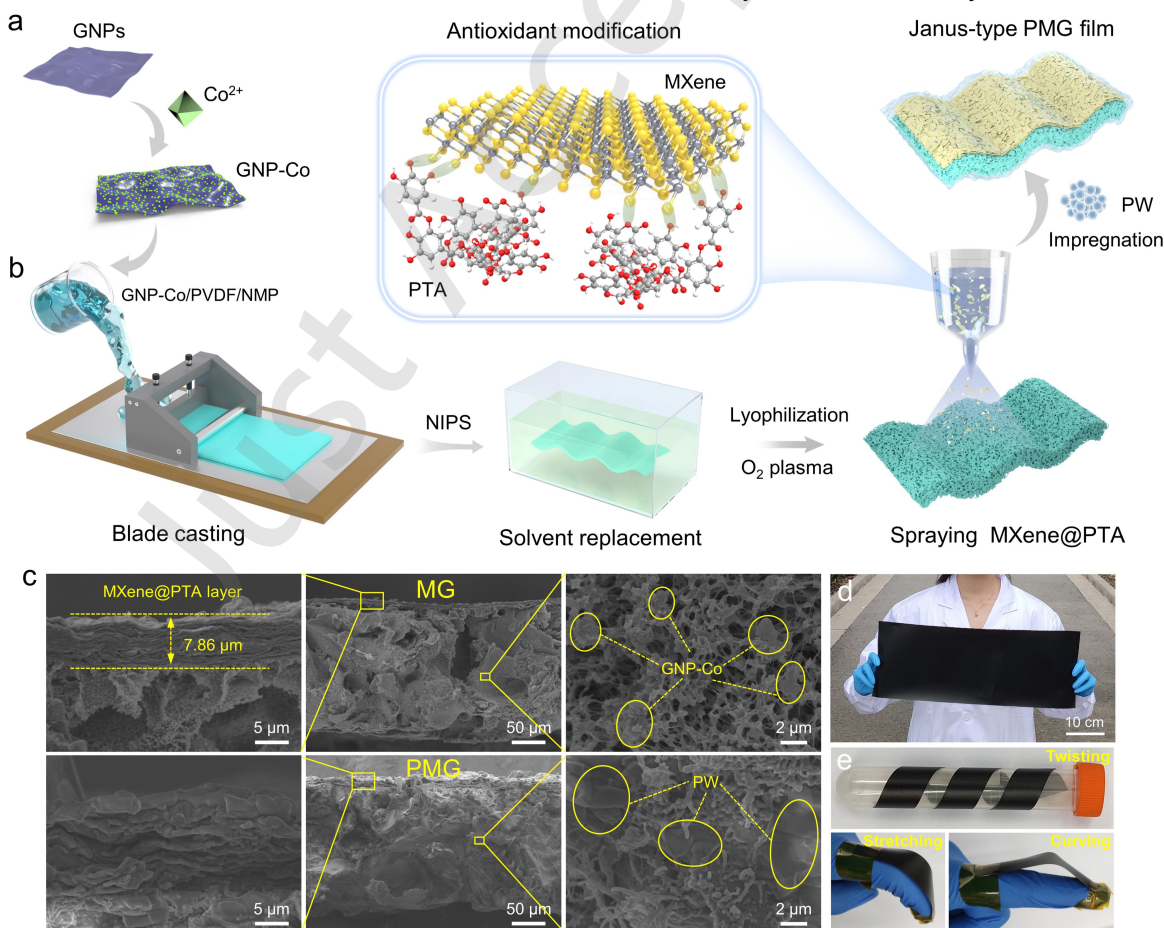


Figure 1 Fabrication and microstructure characterization. (a) Synthesis of heterogeneous GNP-Co hybrids. (b) Scheme depicting the fabrication of Janus-type flexible PMG film using a feasible phase inversion and direct spraying strategy. (c) Cross-sectional SEM images of MG and PMG films. (d) Digital photos of scalable PMG film. (e) Excellent flexibility features of PMG film.

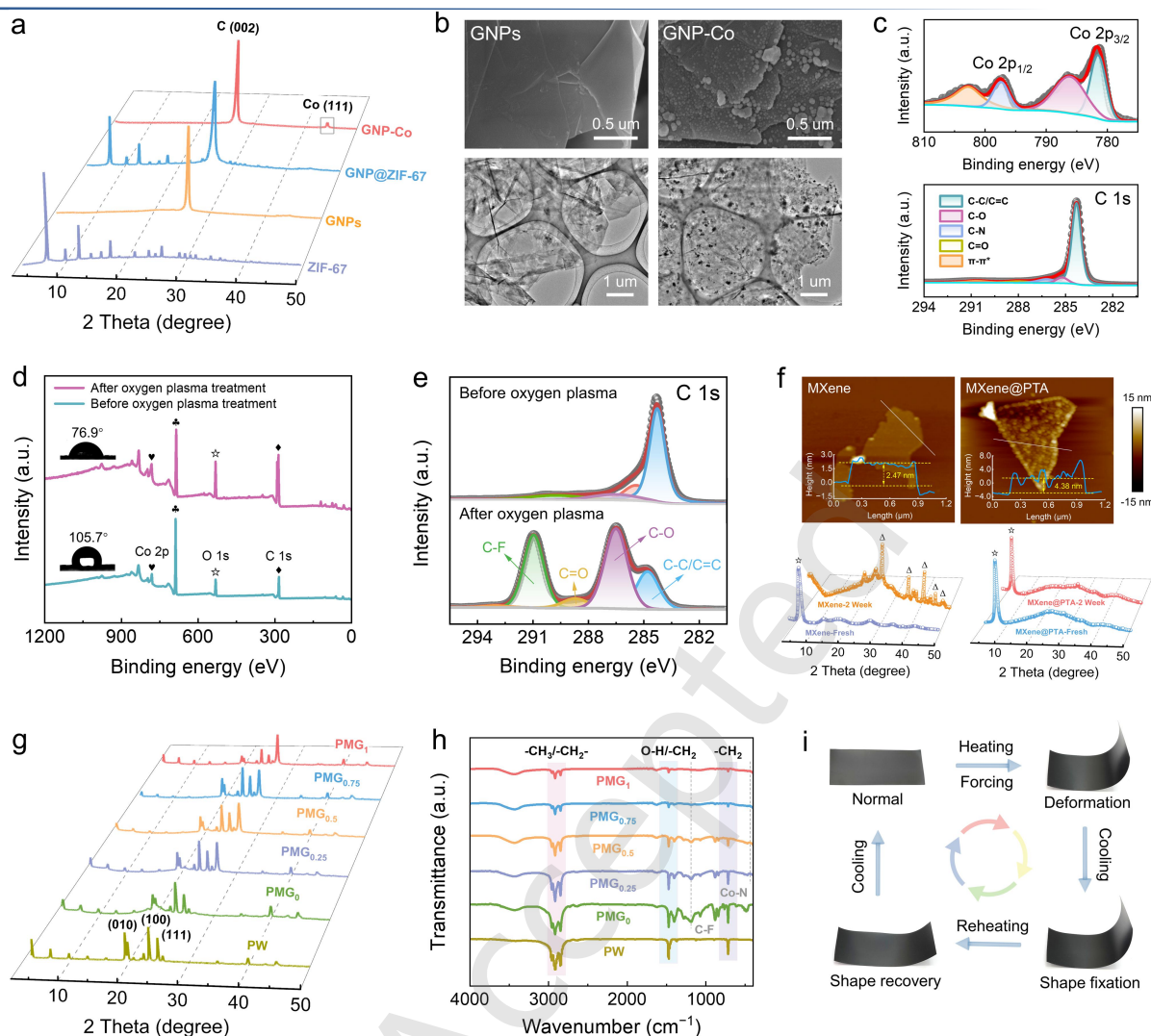


Figure 2 Chemical bond and structure characterization. (a) XRD patterns of ZIF-67, GNPs, GNP@ZIF-67, and GNP-Co. (b) SEM and TEM images of GNPs and GNP-Co. (c) XPS high-resolution Co 2p and C 1s of GNP-Co. (d) XPS survey (inset is the contact angle data) and (e) the corresponding high-resolution C 1s of GNP-Co@PVDF skeleton before and after oxygen plasma treatment. (f) AFM images and XRD patterns of MXene and MXene@PTA. (g) XRD patterns and (h) FT-IR spectra of PW and PMG_x films. (i) Thermal-actuated shape recovery performance of PMG film.

To enhance interfacial interactions, oxygen plasma treatment is used for introducing a plethora of hydrogen bonds at the interface between GNP-Co@PVDF skeleton and MXene@PTA. The treatment results in the introduction of abundant oxygen-containing functional groups, as evidenced by the XPS of GNP-Co@PVDF skeleton (**Fig. 2d, e**). Besides, the plasma treatment significantly reduces the contact angle of GNP-Co@PVDF skeleton from 105.7° to 76.9°, suggesting the successful introduction of hydroxyl groups. Given the propensity of MXene for oxidation, TA is selected to modify MXene for its exceptional adhesive property. The modified MXene@PTA exhibits characteristic peaks of MXene while also appearing new peaks at 1461 cm⁻¹ (**Fig. S2**), corresponding to the stretching vibration of PTA [45]. The thicknesses increase from 2.47 nm for pristine MXene to 4.38 nm for MXene@PTA (**Fig. 2f**) further substantiating the PTA deposition. The hydroxylation of GNP-Co@PVDF skeleton surface is designed to react with the hydroxyl groups of MXene@PTA, facilitating the formation of a hydrogen-bonded MG_x film. As shown in **Fig. 1c**, the sprayed MXene@PTA is assembled into a continuous layer (~7.86 μm) that tightly adheres to G_x film to form a janus structure.

The GNP-Co@PVDF skeleton exhibits a consecutive interpenetrating pore morphology, with pore size distribution between 0.5 ~ 3.0 μm and porosity of 70% ~ 85%, which is highly positive for adsorbing and encapsulating PW. Compared with pure PVDF film prepared by NIPS strategy (**Fig. S3**), the GNP-Co@PVDF skeleton is uniformly distributed and bonded with stacked GNP-Co fillers. The embedded GNP-Co forms consecutive heat-transfer pathways and multi-heterointerfaces for microwave absorption, which enhances the κ and EMI SE of the subsequent PG_x film. The strong capillary effect ensures effective encapsulation of PW into the voids of GNP-Co@PVDF skeleton (**Fig. 1c**). The porous structure of GNP-Co@PVDF skeleton is preserved even after impregnating PW, owing to the moderate impregnation process and excellent mechanical strength of GNP-Co@PVDF skeleton. More importantly, there occurs no interface debonding between GNP-Co@PVDF skeleton and PW matrix, reflecting their commendable compatibility the absence of inner defects or holes. Furthermore, the scalable PMG_x specimen exhibits diverse deformations without fracturing, such as bending, stretching, curving, and twisting (**Fig. 1d, e**). Especially, the

tensile strength and elongation at break of PMG₁ are 2.05 MPa and 39.8%, respectively (**Fig. S4**). This exceptional flexibility renders PMG_x film a good contact with controlled device surface, significantly reducing ultrahigh contact thermal resistance incurred by the brittleness and rigidity of conventional PCMs.

The crystallographic structures of PMG_x films are confirmed by XRD patterns (**Fig. 2g**). The sharp and intense diffraction peaks at $2\theta = 19.3, 23.3,$ and 24.8° are attributed to the monoclinic PW. All characteristic peaks of PW are observable in PMG_x films, without any shift of the peak positions, indicating the crystallization behaviors of PW remain unaltered by incorporation of GNP-Co@PVDF skeleton and MXene@PTA layer. Further information on the interaction and composition of PW, GNP-Co@PVDF skeleton, and MXene@PTA layer are detected by FT-IR spectra (**Fig. 2h**). The PW-specific peaks at 717, 1470, 2850, and 2919 cm^{-1} are associated with the C-H in-plane bending, C-H deformation, $-\text{CH}_2$ asymmetric stretching, and $-\text{CH}_3$ symmetric stretching, respectively. For the PMG_x films, additional peaks at 435 cm^{-1} and 1189 cm^{-1} are indicative of the rocking vibration of Co-N and stretching vibration absorption of C-F, which are characteristic features of the Co and PVDF components, respectively. Note that all intrinsic characteristic peaks of PW, GNP-Co@PVDF skeleton and MXene@PTA layer are visible in the PMG_x films, with the absence of new signals and the lack of displacement in feature bands. The results confirm that the PW encapsulation within the PMG_x films is achieved through physical bonding rather than chemical reaction, aligning with the XRD findings. Moreover, the PMG_x film manifests thermally-actuated shape recovery performance (**Fig. 2i**). The PMG_x spline undergoes deformation when subjected to external force and heated above its phase-transition temperature. The deformation becomes fixed as the spline solidifies. Upon subsequent reheating, the PMG_x spline can spontaneously revert to its original shape. This dynamic thermomechanical property endows the PMG_x film with application potential for the realm of wearable thermal management and smart portable electronics.

3.2 Thermal analysis and phase-change behaviors

Low κ of the PCMs is a longstanding bottleneck hindering the advancement of rapid thermal energy storage and discharge applications [46]. By incorporating GNP-Co hybrids with high κ , the consecutive interconnected GNP-Co@PVDF skeleton is anticipated to be heat-transfer highways for fabricating high-performance PCCs. The κ values of different samples were systematically investigated based on hot disk technique (**Fig. 3a**). As a baseline, raw PW and PMG₀ display a low κ of 0.2 and 0.195 $\text{W m}^{-1} \text{K}^{-1}$, which attributes to impeded phonon vibration arising from the chaotically entangled chains. After introducing GNP-Co@PVDF skeleton, the PMG_x samples exhibit a substantial κ improvement with the increased GNP-Co loading. Notably, the κ value of PMG_x sample is markedly enhanced to 2.12 $\text{W m}^{-1} \text{K}^{-1}$, as the GNP-Co proportion in GNP-Co@PVDF skeleton increases from 20 to 50 wt%. The PMG₁ achieves the highest κ , which is almost 9.6-fold than PW, due to the formation of continuous thermal conduction paths at higher GNP-Co contents. In addition, PMG_x exhibits a slightly higher κ value

compared to that of PG_x at the same loading fraction of GNP-Co (**Fig. S5**), arising from the dense and continuous MXene@PTA layer for rapid heat transport. Collectively, these results underscore that the incorporation of GNP-Co@PVDF skeleton and MXene@PTA layer promotes efficient phonon propagation, mitigating the insulating effect of PW and synergistically enhancing the κ of PCCs. To intuitively assess the heat-transfer performance of PCCs in actual scenarios, three different samples of identical dimensions were subjected to heating via a laser point source (**Fig. 3b**). The temperature profiles were monitored using an IR camera, and the heat flow distributions of the specimens at different heating-up time are given in **Fig. 3c**. The center temperature of PW escalates sharply to 96.9 $^\circ\text{C}$ within 60 s, surpassing that of PMG_{0.5} and PMG₁ by 45.3 and 39.5 $^\circ\text{C}$, because of the local thermal accumulation resulted from the intrinsically low κ . The pronounced difference in heat flux between PMG_{0.5} and PMG₁ is ascribed to the superior formation and increased network density of GNP-Co@PVDF skeleton.

Apart from the κ value that determines thermal storage/discharging velocity, the thermal energy storage capability of PCCs is predominantly governed by their enthalpy and phase-change temperature. As depicted in **Fig. 3d, e**, pure PW shows a melting/solidifying enthalpy ($\Delta H_m/\Delta H_c$) of 249.8/252.2 J g^{-1} , with fusion/crystallization point (T_m/T_c) of 45.05/36.44 $^\circ\text{C}$. The DSC curve of PW reveals a double peak, due to the intermediate “rotator phase” that forms during the transition from ordered crystalline to disordered liquid phase [47]. PMG_x films display similar endothermic/exothermic characteristics to pure PW. Specifically, the peak at $\sim 45^\circ\text{C}$ denotes the solid-liquid phase-transition, while $\sim 39^\circ\text{C}$ peak corresponds to reverse liquid-solid phase-transition. In this temperature interval, GNP-Co@PVDF skeleton and MXene@PTA layer have no thermal phase-transition, and the enthalpy of PMG_x is derived entirely from PW component. Resultantly, the $\Delta H_m/\Delta H_c$ of PMG_x films are lower than that of PW, which are 161.3/162.7, 158.7/160.9, 157.1/157.4, 151.2/152.5, and 146.9/147.2, respectively. Nonetheless, these values significantly surpass the industrial needs of 100 J g^{-1} [36], ensuring sufficient energy density for thermal energy storage. In addition, the difference between T_m and T_c is defined as supercooling degree (ΔT_{sc}). A higher ΔT_{sc} implies a broader temperature span available for heat storage, which can potentially diminish the thermal storage efficiency of the material. As presented in **Fig. 3e**, the PMG_x films exhibit a reduced ΔT_{sc} compared to PW, indicating a more constant operating temperature in practical utilization. The following reasons cause above suppressed subcooling behavior: (i) The incorporation of GNP-Co@PVDF skeleton and MXene@PTA layer into PMG_x composites synergistically enhances the thermo-responsiveness to temperature variation. This signifies that more thermal energy are absorbed/stored, thus accelerating their phase transitions and reducing the T_m . (ii) The capillary forces of porous GNP-Co@PVDF skeleton inhibit the nucleation and movement of PW chains, which impedes the retardation and arrangement of crystal growth on the whole, causing an increased T_c . Besides, the thermal effusivity (e) is a critical indicator in

assessing overall heat-exchange ability with the environment ($e = \sqrt{\rho\kappa\Delta H}$, where κ , ρ and ΔH refer to density, thermal conductivity and enthalpy of PCCs, respectively). Benefited from fast heat conduction speed ($2.12 \text{ W m}^{-1} \text{ K}^{-1}$) and high-level enthalpy (146.9 J g^{-1}), the

PMG₁ manifests a prominent e value of $19.5 \text{ J cm}^{-1.5}(\text{msK})^{-0.5}$ (**Fig. S6**), which notably surpasses that of pure PW ($6.6 \text{ J cm}^{-1.5}(\text{msK})^{-0.5}$). This favorable e value endows PMG₁ with an attractive heat exchange performance.

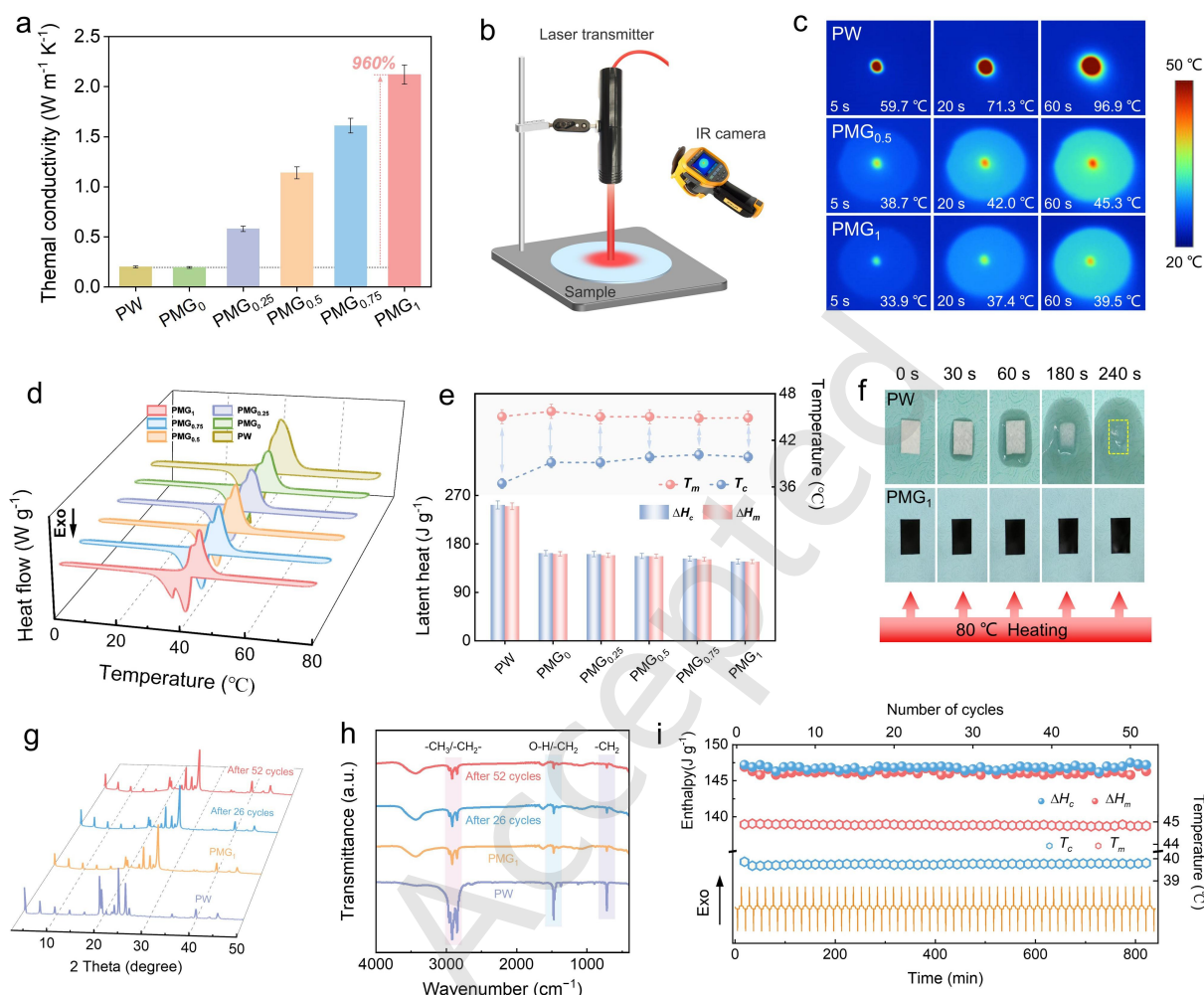


Figure 3 Thermophysical and phase-change properties. (a) Thermal conductivity of PMG_x films. (b) Configuration of laser-IR thermal imaging system and (c) the corresponding IR images showing temperature contours. (d) DSC melting-solidification curves, (e) enthalpy density and subcooling degree of PMG_x films. (f) Photos showing the anti-leakage performance of PMG₁. (g) XRD patterns, and (h) FT-IR spectra of PMG₁ after 52 melting-solidification cycles. (i) enthalpy density, phase-change temperature, and exothermic/endothermic traits of PMG₁ during 52 melting-solidifying cycles.

In actual operating applications, the integrity of PCCs in terms of leakage resistance and shape stability is paramount for safe thermal energy harvesting. As exhibited in **Fig. 3f**, both PW and PMG₁ were tested on a hot-stage to quantify the leakage resistance. At room temperature, both PW and PMG₁ maintain their regular shape. However, upon heating to $80 \text{ }^\circ\text{C}$ ($> T_m$ of PW), the pure PW completely liquidizes and moistens the blotter paper, whereas the PMG₁ sample exhibits no leakage and retains its shape integrity when undergoing extended phase-transition treatment. To further confirm leakage stability, we test the mass-time changes of samples during heating at $80 \text{ }^\circ\text{C}$. Pure PW exhibits a rapid mass loss to just 1% of its initial value within 240 s, whereas PMG₁ retains more than 99% of its original mass under the same conditions. The salient form stability and leakage resistance of PMG₁ stems from following aspects: (i) The molten PW is fully adsorbed and encapsulated within the pores of GNP-Co@PVDF skeleton through sufficient capillary force and huge surface tension. (ii) The MXene@PTA layer

provides additional surface tension forces to further inhibit the fluidity of molten PW chain. (iii) The GNP-Co@PVDF skeleton and MXene@PTA layer form a robust 3D-interconnected supporting scaffold that collectively resists deformation and improves structural stability of PMG₁. Moreover, thermal cycling integrity of PMG₁ was tested to assess its long-term durability and economic feasibility. In **Fig. 3g and h**, the XRD and FT-IR spectra after 52 cycles almost overlap with the original curve, with negligible deviations in the position, intensity and shape of characteristic peaks. This consistency indicates excellent thermal cycling reversibility in both crystal structures and functional groups. Likewise, the endothermic/exothermic peaks of PMG₁ remain consistent even after multiple melting-solidification cycles, exhibiting nearly identical $\Delta H_m/\Delta H_c$ and T_m/T_c without significant attenuation (**Fig. 3i and Table S2**). The results jointly verify the substantial robustness and reversibility of PMG₁ under repeated thermal shocks.

3.3 Superior EMI shielding performance

Desirable EMI-shielding function is essential to mitigate EMWs pollution in advanced electronics, modern military and civil fields. According to Simon formalism, the total EMI shielding effectiveness (SE_T) increases with the electrical conductivity (σ) and thickness (t) [48]:

$$SE_T = 50 + 10 \log(\sigma/f) + 1.7t(\sigma f)^{0.5} \quad (1)$$

where f is frequency of EMWs. As a control, electrically insulating PG₀ exhibits a poor σ of $9.04 \times 10^{-12} \text{ S m}^{-1}$ (Fig. 4a). However, PMG_x films have already transited into conductive substances, showing remarkably increased σ values ranging

from 35.8 to 302.4 S m^{-1} , which vastly surpasses the fundamental requirement of conductive EMI-shielding materials (1 S m^{-1}) [49]. The σ is substantially enhanced by more than 13 orders of magnitude, which stems from the consecutive GNP-Co@PVDF skeleton and MXene@PTA layer for rapid electron transport. In general, high σ is prone to elevate the complex permittivity and consequently strengthen the EMWs absorptivity of PMG_x films [50]. When employing PG₀ and PMG₁ as conductors in a circuit, the brightness of LED visually reflects the great disparity of their σ values. The PMG₁ endows LED display high brightness, while pure PW fails to drive the LED (Fig. S7).

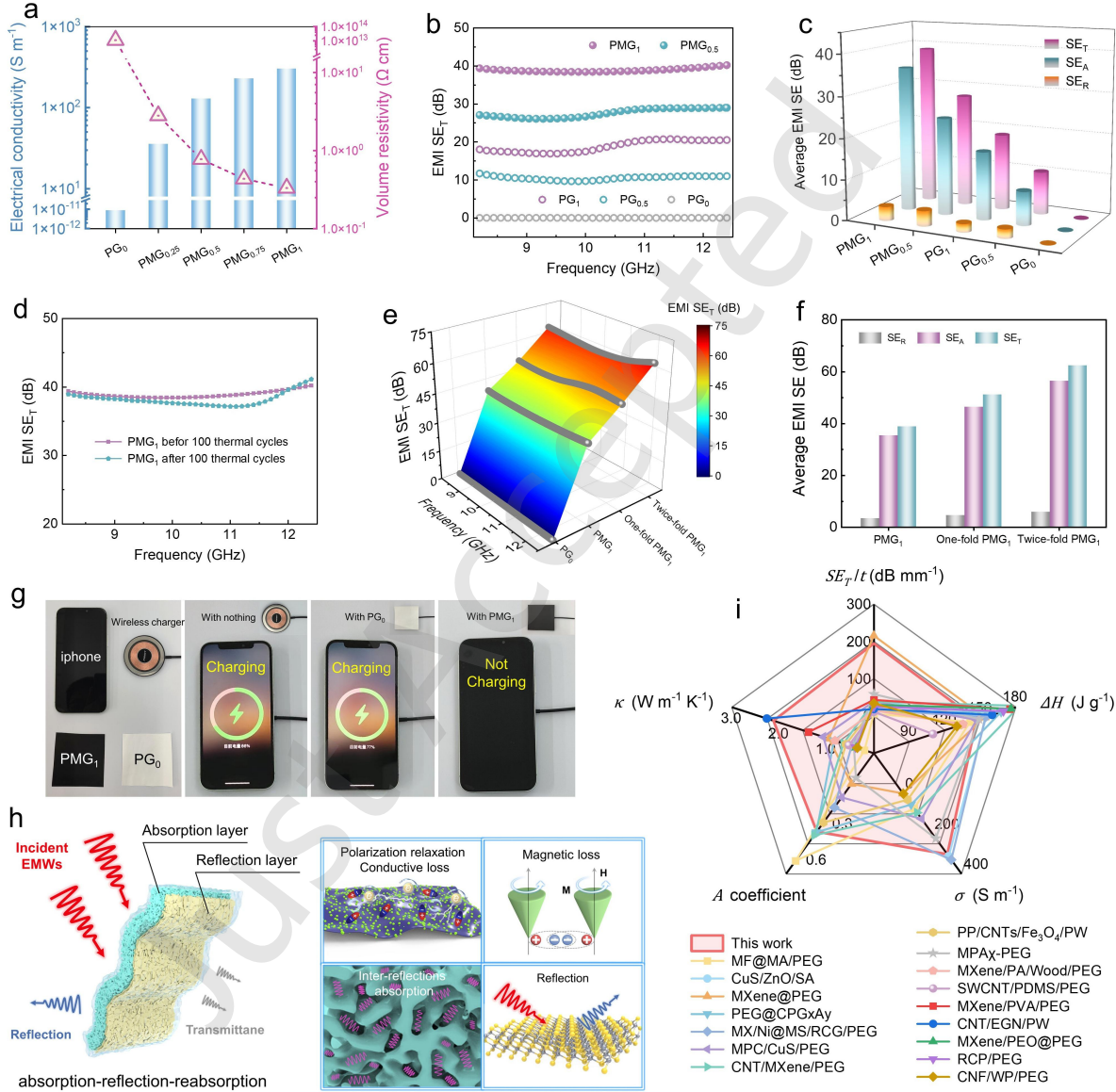


Figure 4 Superior EMI-shielding performance. (a) Electrical conductivity of PMG_x films. (b) Frequency-dependent SE_T , and (c) average SE_T , SE_A , and SE_R of PMG_x films. (d) Frequency-dependent SE_T of PMG₁ before and after 100 thermal cycles. (e) Frequency-dependent SE_T and (f) average SE_T , SE_A , and SE_R of PMG₁ with different thickness. (g) Photos of the signal blocking test between smartphone and wireless charger, showing the practical EMI-shielding ability of PMG₁ film. (h) Scheme illustrating the EMWs attenuation mechanism of the tailored PMG_x films. (i) Comparison of the κ , SE_T/t , ΔH , σ and A with the latest reported PCCs.

The SE_T of various specimens is detected at X-band (8.2-12.4 GHz), which is extensively utilized in wireless communications, wearable electronics, aerospace and military applications. Here, η_{SE} denotes the percentage of incident EMWs blocked by the shielding material and is determined according to equation 2:

$$\eta_{SE} (\%) = 100 - 100 \left(10^{\frac{SE_T}{10}} \right)^{-1} \quad (2)$$

As seen from Fig. 4b, the PG₀ film is nearly transparent to EMWs, showing a terrible $SE_T < 0.025 \text{ dB}$ and extremely low $\eta_{SE} < 0.57\%$. In contrast, the PMG₁ behaves a largely enhanced SE_T of 38.9 dB and η_{SE} of 99.98%, surpassing the primary requirement (20 dB, 99%) of conventional EMI-

shielding material. The SE_T of PMG₁ is markedly superior to that of PMG_{0.5}, attributable to the increased GNP-Co content for forming a well-developed conductive GNP-Co@PVDF skeleton. Meanwhile, the Janus-type PMG_x films manifest better SE_T than PG_x counterparts, this is because MXene@PTA layer can act as an efficacious reflective barrier to attenuate EMWs via a distinctive “absorption-reflection-reabsorption” action. To elucidate the intrinsic EMI-shielding mechanisms, the A - R power coefficients of janus-type PMG_x films are presented (Fig. S8). The PMG_x films exhibit a relatively high A coefficient (~ 0.5) before incident EMWs enter into the interior, indicative of the excellent microwave absorption capacity of heterogeneous GNP-Co hybrids. Moreover, Fig. 4c illustrates the discrete contribution of microwave reflection (SE_R) and microwave absorption (SE_A) to SE_T for janus-type PMG_x films ($SE_T = SE_R + SE_A$, details are available in Supporting Information). All samples show a significantly higher SE_A than SE_R , suggesting that the majority of surviving EMWs within PMG_x films are attenuated by absorption, thereby substantially curtailing the secondary EMWs pollution. For PMG_x films, both SE_T and SE_A exhibit a pronounced increment with escalating GNP-Co content, while SE_R is progressively reduced. This result derives from the synergism of elevated electric/magnetic loss by heterogeneous GNP-Co hybrids. More importantly, the SE_T of PMG₁ film is basically stable without significant attenuation (Fig. 4d), even after enduring 100 thermal cycles. The remarkable EMI-shielding stability enables its huge potential in long-term service. Additionally, Other critical properties such as enthalpy, thermal conductivity, and mechanical response after multiple cycles also validate the long-term durability of PMG₁ (Figure S9).

Given that the SE_T is strongly influenced by material thicknesses, the PMG₁ film were folded once and twice by origami method to evaluate the impact on its EMI-shielding performance. Notably, the SE_T of twice-folded PMG₁ film can be further upgraded from 38.9 to 62.5 dB (Fig. 4e), indicating that the transmission of incident EMWs through PMG₁ is only 0.000056%, with a blocking efficiency over 99.99994%. This outcome verifies the high anti-permeability capability of the folded PMG₁ film in obstructing EMW transmission. The average SE_T of single-layer, once-folded and twice-folded PMG₁ films are 38.9, 51.2 and 62.5 dB, respectively (Fig. 4f). The SE_A accounts for 91.0%, 90.8% and 90.3% of their SE_T , respectively, suggesting that the origami process retains the absorption-dominant principle of EMI shielding. An elevated SE_A can weaken the secondary EMWs pollution induced by microwave reflection. Moreover, the tremendous enhancement on SE_T for the folded PMG₁ film is resulted by the synchronous increment in SE_R and SE_A . Incident EMWs encounter two reflection processes upon penetrating, due to the PMG₁ film possessing a janus-type structure with two interfaces. The reflection paths are largely extended after folding, thus the folded PMG₁ film features a higher R coefficient compared to single-layer film. The R coefficient of twice-folded PMG₁ film is 0.75, beyond 1.36-fold than single-layer film (0.55), indicating that EMWs are converted into thermal energy through multi-reflection and electric/magnetic loss within the material.

Wireless fast charging technology has revolutionized the convenience of device charging. As displayed in Fig. 4g, PG₀ and PMG₁ films were applied to a wireless charging device to compare their EMI-shielding capabilities in practical applications. When PG₀ is sandwiched between the smartphone and the wireless charger, the charging is successful connected. However, the employment of PMG₁ film can immediately interrupt the charging process, halting the transmission of radio energy. These findings confirm that our PMG₁ film possesses impressive anti-electromagnetic interference to effectively shield against external EMWs. The underlying EMI-shielding mechanism of janus-type PMG_x film is systematically shown in Fig. 4h. Based on transmission line theory, the SE_T of material is closely related to conductive charges, electric/magnetic dipoles, and heterogeneous interfaces. Since PW lacks EMI-shielding function, the primary contribution to EMI-shielding essentially depends on GNP-Co@PVDF skeleton and MXene@PTA layer. Unlike bare GNP nanosheets, the adjacent heterogeneous GNP-Co fillers exhibit a larger specific surface area, which facilitates mutual contact and overlap, thereby creating more pathways for both thermal and electrical conduction. The in-situ growth of Co nanoparticles on GNP surfaces significantly reduces the phonon frequency mismatch between the constituent fillers. The resulting multi-heterointerfaces play a key role in EMW attenuation through a synergistic combination of dielectric and magnetic loss mechanisms. These interfacial couplings facilitate charge reordering and the formation of electric dipoles. Moreover, the accumulation of charge carriers at the abundant heterojunctions significantly enhances the EMW attenuation process, ultimately converting electromagnetic energy into heat. Upon irradiation by incoming EMWs, the GNP-Co@PVDF skeleton utilizes abundant free electrons to absorb and attenuate EMWs through impedance matching and polarization loss, efficiently converting EMWs energy into heat and dissipating via hysteresis loss. On the one hand, the interface/dipole polarization within the conductive GNP-Co@PVDF skeleton induces polarization/ohmic loss and contributes to EMW absorption. The magnetic permeability/loss is significantly enhanced by the ferromagnetic Co particles, which enriches the absorbance, attenuation and dissipation of EMWs due to the exchange/natural resonance and magnetic-dielectric coupling action. On the other hand, the porous GNP-Co@PVDF skeleton provides numerous heterointerfaces that facilitate the multi-reflection, dispersion and reabsorption of EMWs. The extension of transmission paths of EMWs can effectively deplete the residual EMWs through hysteresis loss. When the surviving EMWs permeate across GNP-Co@PVDF skeleton to electrically conductive MXene@PTA layer, a portion of EMWs is transformed into heat energy induced by polarization/ohmic loss of MXene, while the other part is immediately reflected to GNP-Co@PVDF side arising from impedance mismatch. The reflected EMWs are subsequently attenuated through magnetic, scattering, and dielectric losses of GNP-Co. Consequently, the “absorption-reflection-reabsorption” mechanism significantly improves the SE_T and maintains a relatively stable absorption rate. The ingenious janus-type structure design capitalizes on the

highly absorbent magnetic properties of GNP-Co@PVDF skeleton and the highly reflective nature of MXene@PTA layer. The synergistic effect of the bilayer electromagnetic loss results in a significant enhancement of EMW energy dissipation. In short, a flexible Janus-type PCC film integrating excellent κ , admirable SE_T/t with a high A coefficient, and high-level ΔH is successfully fabricated. These multifunctional features render our PMG₁ film superior to most reported PCCs in the literature (**Fig. 4i** and **Table S3**), offering vast adaptability in advanced electronics and intelligent wearable products.

3.4 Emerging multi-scenario thermal energy utilization

3.4.1 Sustainable solar-thermal conversion and storage

Efficient solar-harvesting with black low-reflective PCMs has garnered significant interest in the realm of energy storage systems, benefiting from the potential to alleviate the imbalance between thermal energy supply and consumption demand [51]. Nonetheless, the photothermal conversion capability of pristine PCMs is often inadequate for capturing solar energy. To address this limitation, the

GNP-Co@PVDF skeleton and MXene@PTA layer are compounded with PW to augment its photoabsorption property. As depicted in **Fig. 5a**, pure PW shows an inherent weak photon-capturing feature across the entire solar spectrum, even below 0.2 within the wavelength range of 200–1100 nm. By contrast, PMG_x samples exhibit a broadband and near-complete light absorption coefficient (> 0.9) throughout the UV-vis-NIR spectrum (200–2500 nm), especially the PMG₁ counterpart. The hybrid structure of GNP-Co@PVDF skeleton and MXene@PTA layer synergistically promote the photothermal conversion by serving as photo-excited carriers, thereby accelerating the light-matter interactions. Notably, the photosensitive MXene@PTA layer further intensifies optical absorption and mitigates light scattering/reflection through LSPR action. The reduction in high emissivity value of composite triggers nonradiative relaxation and π -plasmon excitation as electrons transition from resonance valence band to conduction band [52]. The exceptionally robust optical absorption and photo-thermal effect of PMG₁ establish a foundation in efficiently harnessing solar energy.

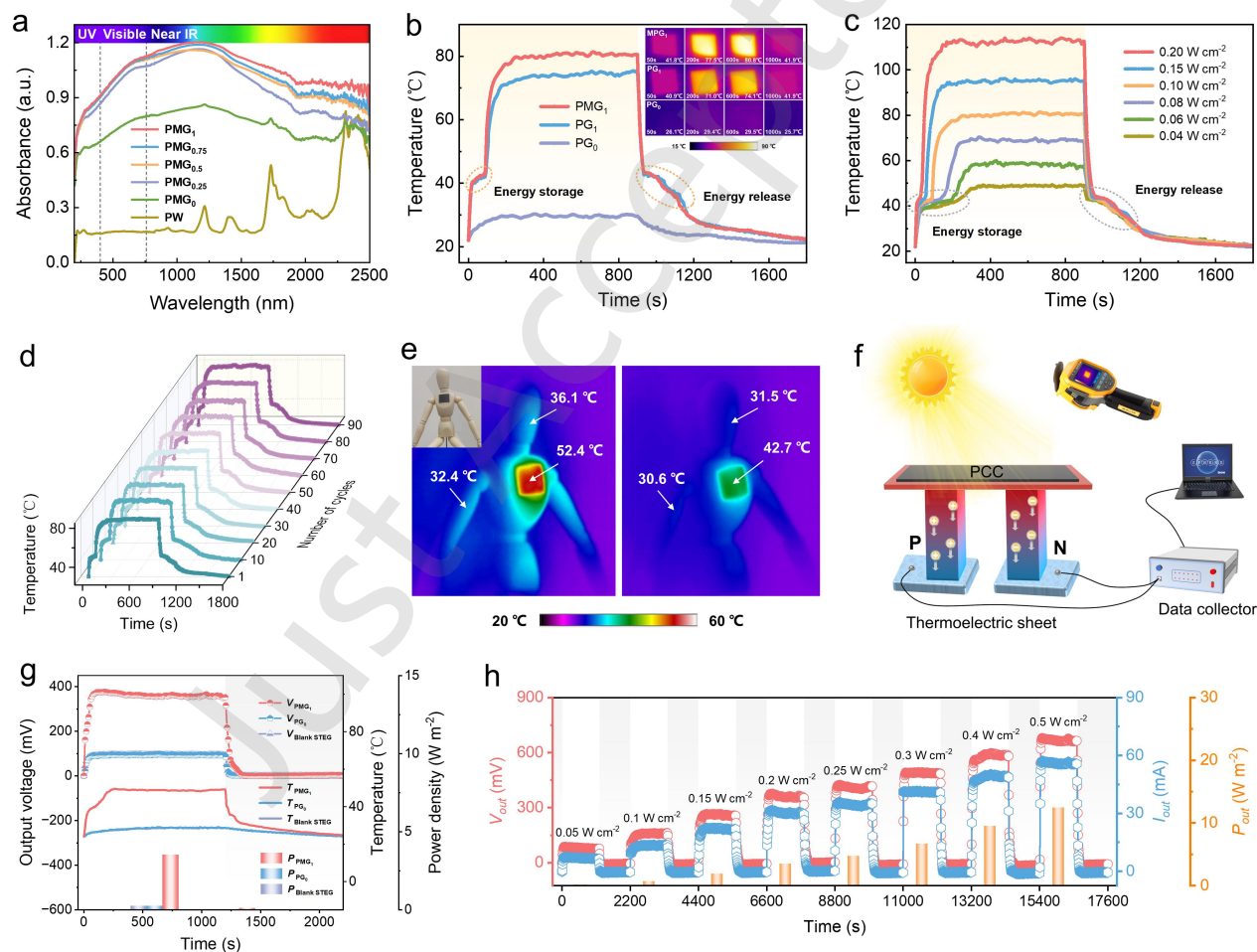


Figure 5 Solar energy utilization based on PMG_x films. (a) UV-vis-NIR absorption spectra of PMG_x films. (b) Temperature-time variations under light irradiation and cessation of irradiation. (c) Photothermal-charging curves of PMG₁ under different solar irradiances. (d) Temperature fluctuations of PMG₁ undergoing 90 thermal cycles. (e) IR images of human model with PMG₁ coated on the chest at irradiation of 236 s and after removing the irradiation for 60 s. (f) Schematic illustration of the STEG device for solar-thermal-electric conversion. (g) Output voltage of blank, PG₀-, and PMG₁-based STEG under illumination intensity of 0.2 W cm⁻². (h) Output voltage/current of PMG₁-based STEG versus the gradient-increased illumination intensity.

To further investigate the photo-heating and energy storage performance, the temperature variations of samples

are real-time monitored by IR camera under the irradiation intensity of 0.10 W cm⁻² (**Fig. 5b**). The PMG₁ sample

demonstrates a significantly accelerated temperature increase compared to PG₁ and PG₀, attributed to the photosensitive MXene@PTA layer and GNP-Co@PVDF skeleton, which efficiently capture the irradiated solar energy and promptly convert it into heat, resulting in temperature increment. The temperature of PMG₁ ascends rapidly to 40.2 °C within 20 s, and later emerges a buffering stage (40.2-42.7 °C) sustained for 66 s, indicating effective solar energy capture and storage as latent heat. Upon the completion of melting phase transition, the temperature of PMG₁ keeps rising to a steady-state of 80.8 °C. Upon removal of the illumination, the temperature of PMG₁ rapidly descends to 43.4 °C, followed by another constant buffering stage (43.4-36.2 °C) sustained for 173 s, invoked by the latent heat release during the solidification. In contrast, PG₀ exhibits an inconspicuous temperature rise within 29.8 °C, without any observable phase transition stages, reflecting its insufficient solar-harvesting capacity under the same irradiation condition. When the light intensity is gradually incremented from 0.04 to 0.2 W cm⁻², the time required to reach phase-transition point is dramatically shortened from 80 to 6 s, and the phase-transition duration of PMG₁ is significantly reduced from 155 to 21 s (**Fig. 5c**). These results confirm that the solar-thermal charging is more easily triggered and accelerated under higher light radiation. Besides, the nearly linear correlation between light intensity and saturation temperature implies precise control over the solar-thermal performance (**Fig. S10**), which is pivotal for practice usage as a light heater. The solar-thermal conversion efficiency (ϕ) of PMG₁ is calculated as 91.2%, according to the following equation:

$$\phi = \frac{m(\Delta H + C_p \Delta T)}{PS\Delta t} \times 100\% \quad (3)$$

where m , ΔH , C_p , and S are the mass, enthalpy, specific heat, and light-receiving area of PMG₁, respectively; ΔT , Δt , and P are the temperature change, phase-transition duration, and solar source, respectively. In contrast, the ϕ of PG₀ is 0 due to its insulating nature. Such excellent ϕ value of PMG₁ can be explained by the following factors: (i) The photon/electron interactions are notably enhanced after incorporating GNP-Co@PVDF skeleton and MXene@PTA layer, which can promote the photon-electron coupling and consequent heat generation through lattice vibrations. (ii) The light-matter interactions are further reinforced through the LSPR action of MXene@PTA layer and GNP-Co fillers [53], accelerating the capture and conversion of sunlight into a non-radiation form of heat energy. Additionally, the PMG₁ displays reversible serviceability and stable reliability during continuous photothermal conversion cycles, with no discernible fluctuations or alterations in the time-temperature progression (**Fig. 5d**). The result highlights the long-term suitability of PMG₁ for repeatable solar-thermal conversion, storage, and discharge.

As mentioned above, the exceptional solar-thermal conversion efficiency and large latent heat of PMG₁ provides the potential for wearable thermal management by providing thermal energy to maintain human thermal comfort under cold conditions. To demonstrate this, the flexible PMG₁ is affixed to the chest of a human model, with the corresponding IR images presented in **Fig. 5e**. Under an incident irradiance of 0.10 W cm⁻², the temperature at the

model's chest region increases to 52.4 °C within 236 s, markedly surpassing the temperature of 32.4 °C observed in the rest of the model not covered by PMG₁. The result demonstrates that the flexible PMG₁ is capable of enabling efficient solar-to-thermal energy conversion, thereby creating a warmer condition for human body. Upon cessation of light radiation, the temperature at the model's chest region is 42.7 °C at 60 s, obviously higher than the rest of the model (30.6 °C) without covering PMG₁. This phenomenon arises from the thermal energy release of PMG₁ to maintain a higher temperature. Above results confirm the significant applicative potential of PMG₁ in the realm of wearable thermal management.

3.4.2 Powerful solar-thermal-electric conversion

Nowadays, photovoltaic cell and solar thermoelectric generator (STEG) represent the predominant techniques for the direct conversion of sustainable sunlight into electricity energy. The Seebeck effect-based STEG has garnered significant interest due to its requirement solely for a temperature differential to generate power [54]. Especially, the integration of STEG with PCMs offers a potential to overcome the space-/time-discrepancy of sunlight. Combining with efficient solar-thermal conversion, rapid thermal storage, large enthalpy density, and anti-EMI radiation performance, our PMG₁ emerges as an undoubtedly value-added component of STEG. Proof as a concept, a prototype STEG equipment was assembled (**Fig. 5f**), comprising Xe lamp as solar simulator, tailored PMG₁ for sunlight harvesting, and commercial thermoelectric module (TGM-241-1.4-1.5, 55 × 55 mm) with a heat sink for cooling. As a contrast, the power-generation-sheet without PMG₁ coverage is utilized as the blank one. At solar irradiance of 0.2 W cm⁻², the temperature of PMG₁ rises rapidly and stabilizes at 49 °C, which is a notable increase of 20.3 °C over the PG₀ and blank power-generation-sheet. Benefiting from the greater temperature difference, the PMG₁-integrated STEG generates a significantly higher output voltage (V_{out}) of 358 mV, marking a substantial improvement of approximately 276% and 273% over the PG₀-loaded and blank STEG, respectively. Meanwhile, the I_{out} always changes in synchronization with the V_{out} (**Fig. 11**). The outcome confirms that more micro-thermoelectric units of power-generation-sheet are activated, arising from enhanced solar-harvesting and thermal charging performance of PMG₁. Noteworthy, even after the cessation of illumination, the PMG₁-loaded STEG sustains electricity generation for an extended period of 152 s. This is attributed to the discharge of thermal energy stored in PMG₁, which extends the power generation duration. Whereas, the V_{out} of blank STEG quickly drops to zero without any decay. The prolonged power generation by phase-transition feedback provides buffering capacity against the inconsistency and intermittence of sunlight, effectively smoothing out electricity fluctuations in real scenarios.

In addition, the dispatchable V_{out} of PMG₁-loaded STEG is systematically investigated at different light intensities to evaluate its long-term working constancy (**Fig. 5h**). Observably, the V_{out} of PMG₁-loaded STEG shows a gradient increase with the escalation of light intensity. In all scenarios, the I_{out} always changes in synchronization with the V_{out} . As

the light density is 0.5 W cm^{-2} , the V_{out} of PMG₁-loaded STEG can reach 670.3 mV, beyond 8.3-fold that of 0.05 W cm^{-2} (81.2 mV). Meanwhile, the calculated power density (P_{out}) of PMG₁-loaded STEG is up to 12.45 W m^{-2} under 0.5 W cm^{-2} illumination ($P_{out} = V_{out} * I_{out}/S$, where S is the hot-side area of thermoelectric sheet). Worth mentioning, the V_{out} of PMG₁-loaded STEG significantly surpasses that of the latest reported counterparts (**Table S4**), due to the superior solar-harvesting capability and fast thermal charging performance of PMG₁. These findings corroborate the high applicability of PMG₁ in solar-thermal-electric conversion, offering a viable solution to reconcile the mismatch between consumer demand and electricity supply. Intriguingly, the PMG₁-loaded STEG connected with passive V_{out} booster can light up LEDs, charge smart bracelet, and energize mobile phone (**Fig. S12**). Series-connected STEGs can provide sufficient electrical power for portable electronics, offering a practical energy solution for remote and harsh environments, including plateaus, deserts, and polar regions.

3.4.3 Electric/magnetic-thermal conversion and storage

In addition to solar-thermal conversion, the PMG₁ is also eminently suitable for Joule heating. It efficiently harnesses the energy from electrification, converting the generated Joule heat into latent heat within PCCs for localized warming or insulation. The PMG₁ sample was connected to a direct-current source at its two ends, and its surface temperature changes were monitored via IR thermal imager. **Fig. 6a** exhibits the trend curves of PMG₁ temperature with time at varying external voltages ranging from 1.5~3.0 V. The saturation temperature (T_s) of PMG₁ escalates with the increase of external voltage, due to the augmenting Joule heat power (according to the Joule's law: $Q = U^2 R^{-1} t$, where Q , U , t and R are Joule heat, voltage, conduction time, and resistance, respectively). Meanwhile, an IR thermal imager visually recorded the color maps of PMG₁ saturation temperature under gradient-changed voltages (**Fig. 6b**). In particular, the T_s of PMG₁ at 1.5, 1.8, 2.1, 2.4, 2.7, and 3.0 V corresponds to 27.3, 31.5, 38.7, 50.2, 68.1, and 92.8 °C. Surprisingly, the T_s can reach 92.8 °C within only 94 s at an input voltage of 3.0 V, indicating the exceptional electro-thermal sensitivity and conversion efficiency of PMG₁. Note that the temperature of PMG₁ increases from 39.9 °C to 42.3 °C within 13.8 s, and the temperature decreases from 43.4 °C to 37.2 °C with a maintenance time of 57.9 s. Meanwhile, the heating rate (0.76 °C/s) is 2.05 times than the cooling rate (0.37 °C/s), ascribed to the sensible heat absorption and latent heat release for Joule heat. Worth mentioning, there is a phase-transition duration for 57.9 s that maintains a constant temperature, which offers a suitable operating temperature for wearable electronic products. As depicted in **Fig. 6c**, PMG₁ exhibits an immediate increase or decrease in surface temperature when the driving voltage is gradually increased or reduced, substantially deriving from the low resistance and swift power response to the current flow. To assess the long-term reliability and endurance of PMG₁ as a Joule

heater, cyclic heating/cooling tests were performed at driving voltage of 1.8, 2.4, and 3.0 V, respectively (**Fig. 6d**). The highly consistent cycling temperature profiles confirm the excellent thermal reversibility and heating stability of PMG₁. Consequently, the tailored PMG₁ is capable of providing effective body warming even at low energy inputs, showing its great serviceability in personal thermal management.

The PMG₁ composite embedding magnetic Co nanoparticles also possess a notable magnetic-thermal conversion ability, which enriches the noncontact energy response and receiving mode of PMG₁, expanding its applicability in wireless thermal therapy for the human body. When subjected to alternative magnetic field, the Co nanoparticles encapsulated in PMG₁ serve as response units to convert magnetic energy into thermal energy through eddy current effect, subsequently stored through PW as latent heat. **Fig. 6e** gives the magnetic hysteresis loop of Co, GNP-Co, and PMG₁. Clearly, PMG₁ exhibits lower magnetic coercivity/retentivity, with the saturation magnetization of 2.95 emu g^{-1} , confirming its superparamagnetic feature. The superior magnetic properties of the GNP-Co are evidenced by its ready adsorption by a magnet, underscoring the capacity of ferromagnetic PMG₁ to convert electromagnetic energy into thermal energy via hysteresis. As a concept verification, the PMG₁ is positioned within a magnetic coil, and its corresponding magnetic-thermal conversion curves is displayed in **Fig. 6f**. Upon exposure to an alternative magnetic field (3 A, 1.2 MHz), the temperature of PMG₁ swiftly ascends to 40.2 °C within 20 s, accompanied by a phase-transition plateaus in the temperature profile, confirming the storage of generated thermal energy as latent heat. Once the phase transition is complete, the temperature of PMG₁ resumes increasing. When the magnetic field is removed, the temperature of PMG₁ rapidly declines, undergoing a phase transition and releasing the stored latent heat from 146 to 244 s. Besides, the magnetic field strength and consequently the magnetic heating rate can be enhanced by increasing the current intensity, which shortens the phase-transition charging duration. Each temperature curve displays a sustained plateau exceeding 90 s during the latent heat release process. Remarkably, PMG₁ can maintain a temperature above 37 °C for over 121 s, with only a charge of 7 s, highlighting its exceptional potential for thermal therapy applications. The underlying magnetic-thermal conversion mechanism of PMG₁ is delineated as follows: the molecular magnetic moments are disordered without alternating magnetic field, governed by Brownian and Néel relaxations [55]. In the above state, the adiabatic temperature is low and the magnetic entropy of the system is large. Upon exposure to an alternating magnetic field, the ferromagnetic Co nanoparticles become magnetized, and their magnetic moment is aligned. As a consequence, the adiabatic temperature rises and magnetic entropy reduces, in turn releasing thermal energy to the surroundings via heat exchange.

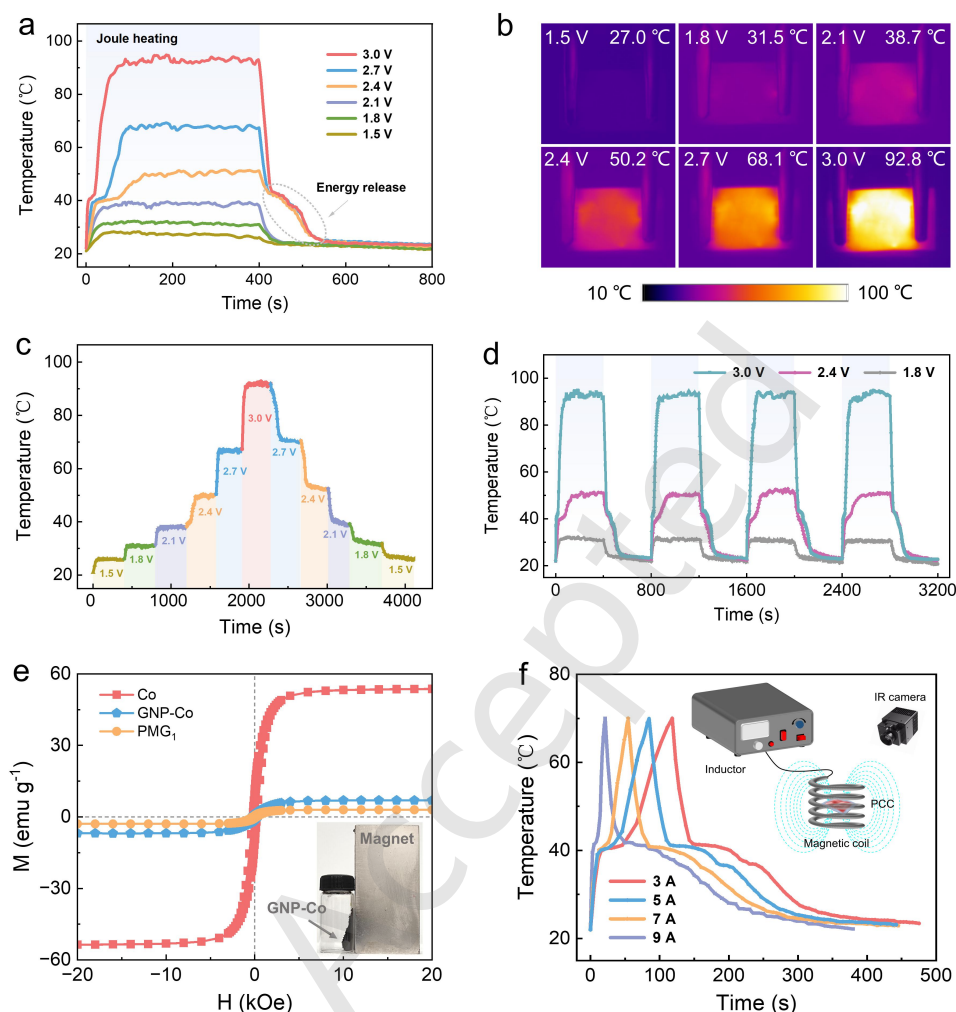


Figure 6 Electric/magnetic-thermal conversion and storage. (a) Surface temperature and (b) the corresponding IR images of PMG₁ at different applied voltages. (c) Temperature evolution under stepwise-increased/decreased voltage adjustment. (d) Electric-thermal cycle test curve at different voltage loads. (e) Magnetic hysteresis loops of Co, GNP-Co, and PMG₁. (f) Magnetic-thermal conversion curves of PMG₁ under different alternating currents.

4 Conclusion

We report the successful synthesis of a novel Janus-type flexible phase change composite film via a facile, cost-effective phase inversion and direct spraying approach, which realizes seamless integration of passive/active thermal management and EMI-shielding functionalities. The composite film incorporates heterogeneous GNP-Co hybrids as thermally conductive and microwave absorption fillers, which are uniformly dispersed in a porous GNP-Co@PVDF skeleton that encapsulates PW. The MXene@PTA layer serves as a highly conductive reflection layer, enabling Janus-type film to attenuate electromagnetic waves via a unique “absorption-reflection-reabsorption” mechanism. The tailored flexible composite film achieves an excellent thermal conductivity of $2.12 \text{ W m}^{-1} \text{ K}^{-1}$ and admirable EMI-shielding effectiveness of 38.9~62.5 dB, beyond 960% and 155, 500% that of pristine PW; together with sufficient enthalpy density of 146.9 J g^{-1} , anti-leakage ability, thermal-driven shape recovery property, and stable cyclic reliability.

Benefiting from exceptional photothermal conversion efficiency of 91.2%, high enthalpy density, and rapid thermal charging rate, the flexible composite film showcases multifaceted energy responsiveness, including efficient photo-thermal/photo-thermal-electric conversion for sustainable solar energy utilization, excellent low-voltage Joule heating ability, and responsive magnetic-thermal conversion performance for contactless thermotherapy. Overall, this study entails new insights into the design and fabrication of multifunctional flexible PCCs for wearable thermal management applications.

Data availability

The data are available upon reasonable request.

Acknowledgements

This work was supported by the China Postdoctoral Science Foundation (2025M780086), the National Key Research and Development Program of China (2024YFD2401300), Tianjin Research Innovation Project for Postgraduate Students

(2023KJ262), and State Grid Corporation of China's Research Program (5419-202019385A). We thank Fanghui Wang from Shijian Lab (www.shijianlab.com) for performing vector network tests.

Declaration of competing interest

The authors declare no competing interests.

Author contribution statement

Hong Guo: conceptualization, investigation, resources, data curation, writing—original draft, visualization. Boyang Hu: methodology, formal analysis, validation, writing—review & editing. Huiting Shan: resources, formal analysis, visualization. Xiao Yang: resources, formal analysis, visualization. Baoan Li: validation, supervision, project administration, funding acquisition. All the authors have approved the final manuscript.

Informed consent

Not applicable.

Ethics statement

Not applicable.

Use of AI statement

None.

References

- [1] Bai, L.; Zhang, S.; Ma, Z.; Dou, L.; Li, J. Bio-inspired multifunctional composite phase change films for synergistic thermal management and microwave absorption. *Nano Res.* **2026**, *19*, 94908229.
- [2] Li, T.; Yang, Y.; Fan, Y.; Huang, D.; Zhang, L.; Hu, X.; Chen, Y.; Sheng, X. Layered MXene-phase change composites for integrated photothermal regulation and electromagnetic shielding. *Nano Res.* **2026**, *19*, 94908054.
- [3] Wang, S.; Ge, J.; Yang, X.; Yu, Y.; He, Q.; Liao, X.; Qin, M.; Feng, W. Bioinspired swelling-deswelling strategy unlocks synergistic molecular solar thermal-fabric systems for personal thermal management. *Adv Mater.* **2025**, *19*, 14043.
- [4] Ciou, Y.-S.; Hsu, C.-K.; Li, J.-W.; Chen, J.-X.; Wang, J.-H.; Chiu, C.-W. Nanocomposite fabrics of GNP/BN/TPU with high thermal conductivity, breathability, stretchability, and excellent comfort for smart cooling wearables. *Adv Compos Hybrid Mater.* **2025**, *8*, 318.
- [5] Yang, P.; Zong, Z.; Wu, Y.; Xia, Z.; Chen, L.; Tang, S. An all-fiber janus film with multi-band synergistic optical modulation for long-term efficient thermal management. *J Mater Sci Technol.* **2026**, *248*, 47-54.
- [6] Zhang, A.; Xiong, Y.; Zhao, Y.; Wu, Y.; Xu, Q.; Ding, Y. A review of passive building thermal management with phase-change materials. *Renewable Sustainable Energy Rev.* **2025**, *211*, 115334.
- [7] Yan, H.; Kou, Y.; Zhang, S.; Liu, H.; Yin, N.; Xu, Y.; Shi, Q. Flexible erythritol-based spatiotemporal phase change films with multiple energy responses for wearable thermal management. *Chem Eng J.* **2025**, *522*, 167630.
- [8] Wang, T.; Wei, R.; Wang, X.; Yang, Y.; Wang, Z.; Wang, Z.; He, Z.; Yi, S. Shape-stabilized and flexible phase change materials with enhanced photothermal conversion for efficient thermal energy storage. *Compos Part B-Eng.* **2025**, *302*, 112539.
- [9] Hu, B.; Guo, H.; Li, T.; Cao, X.; Cao, M.; Qi, W.; Cui, Y.; Li, B. Engineering tiramisu-like phase change nanocomposite for superior thermal energy management and electromagnetic interference shielding. *J Mater Sci Technol.* **2025**, *206*, 113-124.
- [10] Hu, X.; Ai, B.; Quan, B.; Shi, Z.; Liu, Z.; Sheng, M.; Ding, Y.; Yan, H.; Xiao, Y.; Lu, G.; Liu, S.; Wu, H.; Huang, X.; Yan, X.; Lu, X.; Qu, J. Synergistically networked 1D/2D/3D phase change composites with ultra-fast contactless thermal energy conversion and enhanced electromagnetic interference shielding performance. *Chem Eng J.* **2025**, *504*, 158915.
- [11] Liu, Y.; Cao, X.; Yang, M.; Wei, M.; Mao, J.; Sun, J.; Xiang, H.; Wang, H. Flame-retardant wood-based composite phase change materials based on polydopamine functionalized carbon dots for efficient solar-to-thermal energy storage. *Adv Compos Hybrid Mater.* **2025**, *8*, 400.
- [12] Tong, J.; Tao, Z.; Hou, W.; Yang, C.; Yuan, T.; Liao, H.; Huang, X.; Wang, G. Recent advances in the properties, synthesis, and applications of oriented composite phase change materials. *J Energy Chem.* **2026**, *112*, 902-944.
- [13] Cao, Y.; Zhao, Z.; Zeng, X.; Teng, J.; Huang, J.; Min, Y. High-performance polyimide/polypyrrole-CNTs@PEG composites for integrated thermal management and enhanced electromagnetic wave absorption. *Adv Compos Hybrid Mater.* **2025**, *8*, 104.
- [14] Wu, M.; Xuan, Y.; Liu, X.; Jing, Y.; Li, T. Flexible, recyclable, and highly conductive self-healing polymer-based phase change films for thermal management. *Adv Funct Mater.* **2025**, *35*, 2506229.
- [15] Feng, Y.; Hai, G.; Sun, G.; Chen, K.; Wang, X.; Zhao, J.; Li, Y.; Chen, X. Dual-functional phase change composites integrating thermal buffering and electromagnetic wave absorption via multi-interfacial engineering. *Advanced Fiber Materials.* **2025**, *7*, 1873-1887.
- [16] Li, P.; Liu, Z.; Peng, Y.; Yang, S.; Meng, T.; Hu, Y.; Jiang, Y.; Sun, H.; Li, Q.; Hu, X. Fast thermal responsive separators toward long-life and safe lithium metal batteries. *Nano Res.* **2023**, *17*, 2746-2754.
- [17] Liu, W.; Ren, S.; Liu, R.; Li, A.; Du, Q.; Dong, W.; Bai, J.; Chen, X.; Sun, H.; Dong, C. Efficient and stable solar-thermal energy storage via camel-hump-like adsorption strategy using nano-ribbon-modulated ultra-light self-standing film. *Compos Part B-Eng.* **2025**, *298*, 112350.
- [18] Zhang, C.; Mao, Y.; Li, K.; Liu, Y.; Xu, Z.; Pang, K.; Cai, S.; Fan, L.; Gao, C. High power and energy density graphene phase change composite materials for efficient thermal management of Li-ion batteries. *Energy Storage Mater.* **2025**, *75*, 104003.
- [19] Zhang, X.; Sun, K.; Liu, H.; Chen, J.; Yan, X.; Kou, Y.; Shi, Q. Flexible insulating phase change composite film with improved thermal conductivity for wearable thermal management. *Nano Energy.* **2024**, *121*, 109256.
- [20] Chen, Y.; Yu, H.; Feng, Y.; Feng, W. Custom-assembled phase change modular devices for personalize speciality: Carbon energy thermal management application. *Nano Res.* **2025**, *18*, 94907202.
- [21] Kang, Z.; Zhao, J.; Chen, Z.; Liu, X.; He, M. Flame-retardant and phase-changing microcapsules incorporating black phosphorus for efficient solar energy storage. *J Cleaner Prod.* **2024**, *467*, 143055.
- [22] Li, Y.; Wang, X.; Chen, K.; Feng, Y.; Liu, P.; Jiang, Y.; Shen, Z.; Chen, X. Hierarchical MoS₂/CuS photonic nanostructure accelerating photothermoelectric conversion of bacterial cellulose based phase change materials. *J Colloid Interface Sci.* **2026**, *702*, 138863.
- [23] Ye, X.; Yang, D.; Yu, L.; Jiang, P.; Liu, W.; Lou, H. Phase change material composites based on 3D lignin-derived porous carbon prepared by in-situ activation for efficient solar-driven energy conversion and storage. *J Colloid Interface Sci.* **2025**, *678*, 704-719.
- [24] Liang, S.; Wang, X.; Chen, L.; Qiu, X. Utilizing heterogeneity of lignin to diminish supercooling of phase change material nano-capsules with high latent heat. *J Colloid Interface Sci.* **2025**, *683*, 833-840.
- [25] Hou, M.; Chen, H.; Li, S.; Zhang, X.; Chen, J.; Jiang, Z.; Wang, C.; Lai, N. C.; Ding, Y. A tri-mode photothermal, phase-change, and radiative-cooling film for all-day thermoelectric generation. *Adv Mater.* **2025**, *37*, 2505601.
- [26] Das, M.; Ura, D. P.; Szweczyk, P. K.; Berniak, K.; Knapczyk-Korczak, J.; Marzec, M. M.; Pichór, W.; Stachewicz, U. Thermal energy storage

performance of liquid polyethylene glycol in core-shell polycarbonate and reduced graphene oxide fibers. *Adv Compos Hybrid Mater.* **2024**, *7*, 123.

[27] Xiao, Y.; Lin, J.; Cao, Y.; Liu, X.; Wang, H.; Alshammari, S.; Shi, X.; Guo, H.; Helal, M. H.; Algadi, H.; Ren, J.; Zhumanova, N.; Akimbayeva, N.; Toktarbay, Z.; Huang, J.; Guo, Z. Coaxial electrospun PEG-4000@PVDF composite membranes with intumescent flame retardants: Toward multifunctional fabrics for thermal management and ultra-safe lithium-ion battery separators. *J Mater Sci Technol.* **2026**, *251*, 282-296.

[28] Zhang, R.; Li, Y.; Zhou, B.; Liu, C.; Zhang, Y.; Xiao, C.; Liu, Q.; Fu, X.; Hu, S.; Hu, T.; Wong, C. P. Recyclable solid-solid phase change polymer-based composites with high latent heat and ultrahigh thermal conductivity via dynamic internal-catalyzed esterification and vertically oriented conductive pathway. *Small.* **2025**, *21*, 2501959.

[29] Sun, L.; Huang, Y.; Chen, W.; Jiang, X.; Hu, X. Chemically crosslinked flexible polymer phase change material with self-healing, self-adaptive, and shape-memory properties for thermal interface applications. *J Colloid Interface Sci.* **2025**, *695*, 137831.

[30] Chen, X.; Lin, J.; Feng, Y.; Chen, K.; Qin, M.; Han, S.; Jiang, Y.; Shen, Z.; Li, Y. Carbon-metal network boosting photon/phonon transport in photothermal phase change materials. *Carbon* **2025**, *238*, 120192.

[31] Li, S.; Zhang, N.; Chen, H.; Zhang, Z.; Zhao, Y.; Yang, Y.; Shang, H.; Wang, D.; Zhang, B. Encapsulating phase change materials into melamine formaldehyde sponge assembled with polypyrrole modified halloysite nanotube for effective solar-thermal energy storage and solar-thermal-electric conversion. *J Colloid Interface Sci.* **2025**, *682*, 423-435.

[32] Wang, R.; Geng, L.; Zhang, X.; Liang, G.; Zhao, J.; Liu, C. Synthesis of flexible form stable phase change materials with in-situ formed porous TiO₂ for personal thermal management. *Chem Eng J.* **2025**, *505*, 159592.

[33] Zhao, X.; Zou, D.; Wang, S. Flexible phase change materials: Preparation, properties and application. *Chem Eng J.* **2022**, *431*, 134231.

[34] Wang, S.; Du, R.; Li, T. Progress and perspective on thermal conductivity enhancement of phase change materials. *Science Bulletin* **2024**, *69*, 3176-3179.

[35] Hu, B.; Guo, H.; Cui, Y.; Li, J.; Cao, M.; Qi, W.; Cao, X.; Li, B. Engineering multifunctional phase change composites enabled by dual-interpenetrating hybrid scaffold for excellent thermal conductivity and electromagnetic absorption. *Chem Eng J.* **2024**, *492*, 152259.

[36] Hu, B.; Guo, H.; Liu, J.; Cao, M.; Li, M.; Cui, Y.; Li, B. Vertically oriented boron nitride/silicon carbide scaffold for thermal-conductive and electrical-insulating phase change composites. *Compos Part A-Appl S.* **2023**, *168*, 107460.

[37] Liu, L.; Zhang, Y.; Zhang, S.; Tang, B. Advanced phase change materials from natural perspectives: Structural design and functional applications. *Adv Sci.* **2023**, *10*, 2207652.

[38] Gao, Y.; Chen, X.; Jin, X.; Zhang, C.; Zhang, X.; Liu, X.; Li, Y.; Li, Y.; Lin, J.; Gao, H.; Wang, G. Multifunction integration within magnetic CNT-bridged MXene/CoNi based phase change materials. *eScience* **2024**, *4*, 100292.

[39] Hu, L.; Zhang, L.; Cui, W.; An, Q.; Ma, T.; Wang, Q.; Mai, L. Carbon-based porous materials for performance-enhanced composite phase change materials in thermal energy storage: Materials, fabrication and applications. *J Mater Sci Technol.* **2025**, *210*, 204-226.

[40] Li, Y.; Liu, P.; Gao, Y.; Feng, Y.; Li, P.; Chen, X. Scattered co-anchored MoS₂ synergistically boosting photothermal capture and storage of phase change materials. *J Energy Chem.* **2024**, *95*, 208-215.

[41] Li, Y.; Liu, P.; Li, P.; Feng, Y.; Gao, Y.; Diao, X.; Chen, X.; Wang, G. Neural network-inspired hybrid aerogel boosting solar thermal storage and microwave absorption. *Nano Research Energy* **2024**, *3*, 9120120.

[42] Li, X.; Sheng, M.; Gong, S.; Wu, H.; Chen, X.; Lu, X.; Qu, J. Flexible and multifunctional phase change composites featuring high-efficiency

electromagnetic interference shielding and thermal management for use in electronic devices. *Chem Eng J.* **2022**, *430*, 132928.

[43] Zhu, C.; Hao, Y.; Wu, H.; Chen, M.; Quan, B.; Liu, S.; Hu, X.; Liu, S.; Ji, Q.; Lu, X.; Qu, J. Self-assembly of binderless mxene aerogel for multiple-scenario and responsive phase change composites with ultrahigh thermal energy storage density and exceptional electromagnetic interference shielding. *Nano-Micro Letters* **2023**, *16*, 57.

[44] Shahzad, F.; Alhabeib, M.; Christine B. Hatter; Anasori, B.; Hong, S. M.; Koo, C. M.; Gogotsi, Y. Electromagnetic interference shielding with 2D transition metal carbides (MXenes). *Science* **2016**, *353*, 1137-1140.

[45] He, X.; Wu, J.; Huang, X.; Chen, Y.; Zhang, L.; Sheng, X. Three-in-one polymer nanocomposite coating via constructing tannic acid functionalized MXene/BP hybrids with superior corrosion resistance, friction resistance, and flame-retardancy. *Chem Eng Sci.* **2024**, *283*, 119429.

[46] Wu, S.; Li, T.; Tong, Z.; Chao, J.; Zhai, T.; Xu, J.; Yan, T.; Wu, M.; Xu, Z.; Bao, H.; Deng, T.; Wang, R. High-performance thermally conductive phase change composites by large-size oriented graphite sheets for scalable thermal energy harvesting. *Adv Mater.* **2019**, *31*, 1905099.

[47] Guo, H.; Jiao, W.; Jin, H.; Yuan, Z.; He, X. Microsphere structure composite phase change material with anti-leakage, self-sensing, and photothermal conversion properties for thermal energy harvesting and multi-functional sensor. *Adv Funct Mater.* **2022**, *33*, 2209345.

[48] Ma, Z.; Kang, S.; Ma, J.; Shao, L.; Zhang, Y.; Liu, C.; Wei, A.; Xiang, X.; Wei, L.; Gu, J. Ultraflexible and mechanically strong double-layered aramid nanofiber-Ti₃C₂T_x mxene/silver nanowire nanocomposite papers for high-performance electromagnetic interference shielding. *ACS Nano* **2020**, *14*, 8368-8382.

[49] Hu, B.; Guo, H.; Li, T.; Li, J.; Cao, M.; Qi, W.; Li, X.; Li, B. Millefeuille-like phase change composite with multi-level oriented architecture for efficient thermal management and electromagnetic interference shielding. *Chem Eng J.* **2024**, *483*, 149240.

[50] Wang, L.; Ma, Z.; Qiu, H.; Zhang, Y.; Yu, Z.; Gu, J. Significantly enhanced electromagnetic interference shielding performances of epoxy nanocomposites with long-range aligned lamellar structures. *Nano-Micro Letters* **2022**, *14*, 224.

[51] Qiu, C.; Jiang, H.; Liu, P.; Jiang, C.; Ji, X.; Zhang, Y.; Sun, W.; Guo, X. An ultrastrong wood-based phase change material for efficient photothermal conversion and thermal energy conservation. *Compos Part B-Eng.* **2024**, *279*, 111460.

[52] Tang, Z.; Gao, Y.; Cheng, P.; Jiang, Y.; Xu, J.; Chen, X.; Li, A.; Wang, G. Metal-organic framework derived magnetic phase change nanocage for fast-charging solar-thermal energy conversion. *Nano Energy* **2022**, *99*, 107383.

[53] Xu, J.; Jiang, Y.; Chen, X.; Tang, Z.; Gao, Y.; Huang, M.; Wan, Y.; Cheng, P.; Wang, G. Photo- and magneto-responsive highly graphitized carbon based phase change composites for energy conversion and storage. *Materials Today Nano* **2022**, *19*, 100234.

[54] Liu, J.; Du, C.; Li, Z.; Chai, C.; Sun, Q.; Huo, B.; Chen, G. Bioinspired thermo-actuated thermoelectric/phase-change system for all-day power generation. *Adv Compos Hybrid Mater.* **2026**, *9*, 47.

[55] Gao, Y.; Tang, Z.; Chen, X.; Yan, J.; Jiang, Y.; Xu, J.; Tao, Z.; Wang, L.; Liu, Z.; Wang, G. Magnetically accelerated thermal energy storage within Fe₃O₄-anchored MXene-based phase change materials. *Aggregate* **2022**, *4*, e248.

© The Author(s) 2026. *Nano Research* published by Tsinghua University Press. The articles published in this open access journal are distributed under the terms of the Creative Commons Attribution 4.0 International License (<http://creativecommons.org/licenses/by/4.0/>), which permits use, distribution and reproduction in any medium, provided the original work is properly cited.

Janus-type flexible phase change composite for wearable thermal management and electromagnetic interference shielding

Hong Guo ^{1,2}, Boyang Hu ²✉, Huiting Shan ^{3,4}, Xiao Yang ⁵, Baoan Li ²✉

¹ Department of Energy and Chemical Engineering, Tianjin Ren'ai College, Tianjin 301636, China

² Carbon Neutrality Interdisciplinary Science Centre, Tianjin Key Laboratory of Environmental Technology for Complex Trans-media Pollution, College of Environmental Science and Engineering, Nankai University, Tianjin 300350, China

³ Department of Chemical and Biomolecular Engineering, National University of Singapore, Singapore 117585, Singapore

⁴ Joint School of National University of Singapore and Tianjin University, International Campus of Tianjin University, Fuzhou 350207, China

⁵ Hangzhou Huadian Huayuan Environment Engineering Co., Ltd., Hangzhou 310030, China

Supporting information to <https://doi.org/10.26599/NR.2026.94909156>

Just Accepted

Supplementary Texts

Section S1 Materials and Chemicals

Graphene nanoplates (GNPs, XF022-1) and Ti_3AlC_2 MAX phase powders (XFK03) were obtained from Nanjing XFNANO Materials Tech Co., Ltd. Cobalt nitrate hexahydrate ($\text{Co}(\text{NO}_3)_2 \cdot 6\text{H}_2\text{O}$), 2-MeIm and methanol were purchased from Sinopharm Chemical Reagent Co., Ltd. Polyvinylidene fluoride (PVDF, 99%) was supplied by Beijing HWRKChem Co., Ltd. N-methyl pyrrolidone (NMP, 99%), Lithium fluoride (LiF), and hydrochloric acid were bought from Shanghai Aladdin Biochemical Technology Co., Ltd. Paraffin wax (PW, Brand for OP44E) was supplied by Ruhr Energy Technology Co., Ltd. All chemicals were used as received without further purification.

Section S2 Characterizations

The microscopic structures of nanofillers were investigated by high-resolution transmission electron microscopy (TEM, JEM 2100F, Japan) and atomic force microscopy (AFM, Bruker Dimension ICON, Germany). The morphology of porous skeleton and PCCs was observed using field-emission scanning electron microscopy (SEM, S4800, Hitachi, Japan) with an accelerating voltage of 3 kV. Crystalline structures were examined using X-ray diffraction analysis (XRD, Advance D8, Bruker, Germany) with Cu K α radiation ($\lambda = 1.5406 \text{ \AA}$). Fourier-transform infrared spectrometer (FT-IR, Thermo Scientific, USA) and laser Raman spectrometer (Renishaw inVia, UK) were employed to analyze the intermolecular interactions of the samples. The chemical composition was investigated using X-ray photoelectron spectroscopy (XPS, K-Alpha⁺, Thermo Fisher, USA). Phase-change temperatures and latent heat of samples were investigated via a differential scanning calorimetry (DSC, TA Q20, USA) with a heating/cooling rate of $10 \text{ }^\circ\text{C} \cdot \text{min}^{-1}$ under 20 ml min^{-1} nitrogen atmosphere. The measurement uncertainties for the phase transition temperature are within $\pm 0.5 \text{ }^\circ\text{C}$, while those for the phase transition enthalpy are within $\pm 1\%$. Thermal conductivity coefficients (κ) of PCCs was determined using Hot Disk TPS2200 thermal constant analyzer (AB Co., Sweden), according to standard ISO 22007-2:2008. All samples were tested more than three times and the average value was taken as the final measurement result. A four-probe tester (ST2258C, China) was applied to evaluate the electrical conductivity of specimens at room temperature. EMI shielding efficiency of PCCs were monitored using a vector network analyzer (E5071C, Agilent, USA) over a frequency range of 8.2-12.4 GHz (X-band). Photoabsorption spectra of the PCCs were performed on an ultraviolet-visible (UV-vis) spectrometer (Lambda 950, PerkinElmer, USA) over the wavelength of 200-2500 nm. The solar-thermal conversion capability of PCCs was characterized using a power-adjustable Xenon lamp (CHF-XM500). Surface temperature and thermal images were captured using an infrared camera (TiS75⁺, Fluke, USA).

Section S3 The calculation of electromagnetic shielding parameters

The film samples were cut into the rectangular shape with a dimension of $22.9 \times 10.2 \text{ mm}^2$ for measurements. A vector network analyzer (E5071C, Agilent, USA) and waveguides are employed to detect the scattering parameters of films within X-band (8.2–12.4 GHz). Here, the scattering parameters (S_{11} , S_{21} , S_{12} , and S_{22}) are the power balance of EMWs that interact with shielding material [1]. The four subparameters of S_{11} , S_{21} , S_{12} , and S_{22} are forward reflection coefficient, forward transmission coefficient, reverse transmission coefficient, and reverse reflection coefficient, respectively. The related EMI shielding parameters, including power coefficient of reflection (R), absorption (A), and transmission (T), can be obtained by Eqs. S1-S10 [2]:

$$R = |S_{11}|^2 = |S_{22}|^2 \quad (\text{S1})$$

$$T = |S_{21}|^2 = |S_{12}|^2 \quad (\text{S2})$$

$$A = 1 - (R + T) \quad (\text{S3})$$

$$A_{eff} = (1 - R - T)/(1 - R) \quad (\text{S4})$$

where A_{eff} is the effective absorption coefficient.

According to “Schelkunoff theory”, the corresponding SE_R , SE_A , and SE_T are calculated by the equations below [3]:

$$SE_R = -10 \log(1 - R) \quad (\text{S5})$$

$$SE_A = -10 \log[T/(1 - R)] \quad (\text{S6})$$

$$SE_T = -10 \log T \quad (\text{S7})$$

$$SE_T = SE_R + SE_A + SE_M \quad (\text{S8})$$

where SE_T , SE_R , and SE_A represent total EMI shielding effectiveness, reflection loss, and absorption loss, respectively. SE_M is multiple internal reflection value and usually can be negligible at the time of $SE_T \geq 15 \text{ dB}$.

$$\eta_{SE} = \left[1 - \left(10^{\frac{SE_T}{10}} \right)^{-1} \right] \times 100\% \quad (\text{S9})$$

Where η_{SE} is shielding efficiency and means the ability of a material to block EMWs in terms of percentage.

To compare the effectiveness of shielding materials equitably, thickness of the materials is also taken into account. The related equations were described as:

$$SE_T/t = SE_T/\text{thickness} \quad (\text{S10})$$

Supplementary Figures

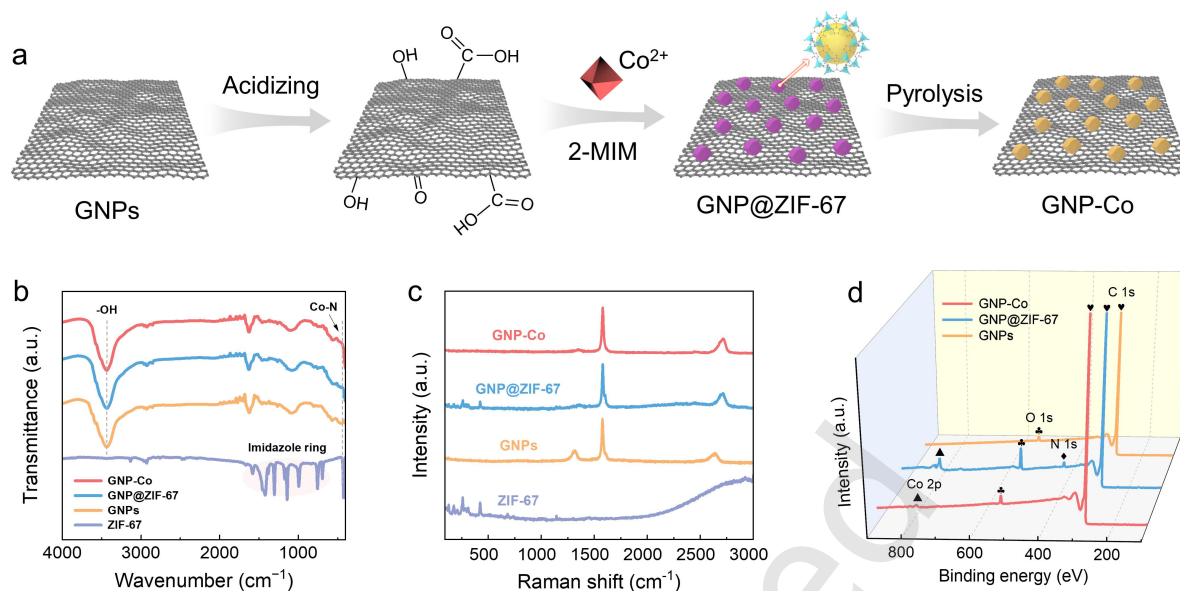


Figure S1 (a) Scheme depicting the synthesis of heterogeneous GNP-Co hybrids. (b) FT-IR spectra and (c) Raman spectra of ZIF-67, GNPs, GNP@ZIF-67, and GNP-Co. (d) XPS survey of GNPs, GNP@ZIF-67, and GNP-Co.

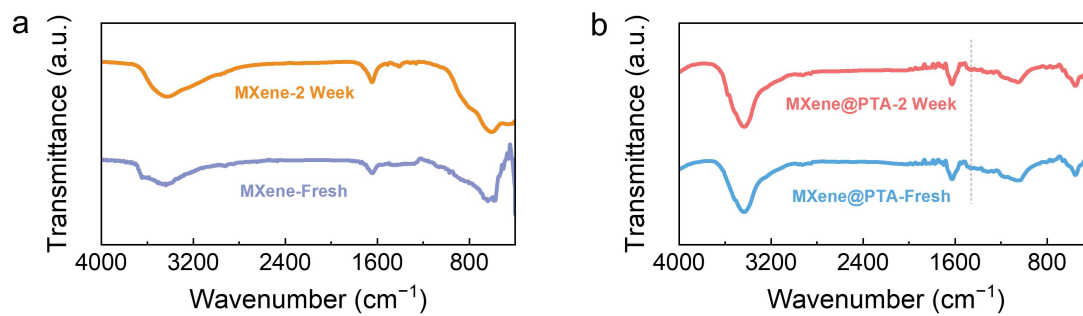


Figure S2 FT-IR spectra of MXene and MXene@PTA, indicating their antioxidant capacity.

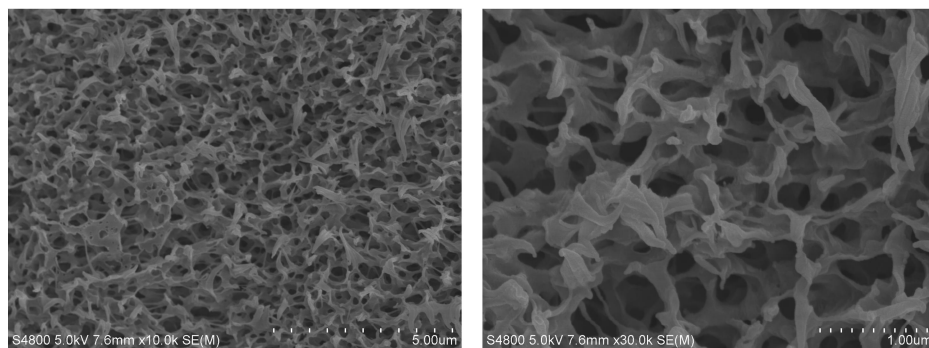


Figure S3 Cross-sectional SEM images of pure PVDF film prepared by NIPS strategy.

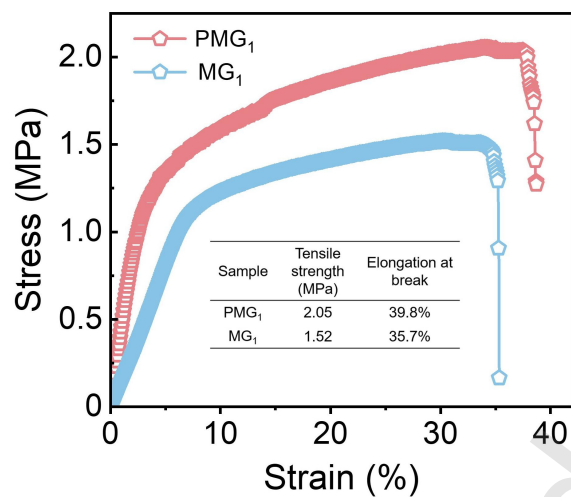


Figure S4 The tensile stress-strain curves of MG₁ and PMG₁.

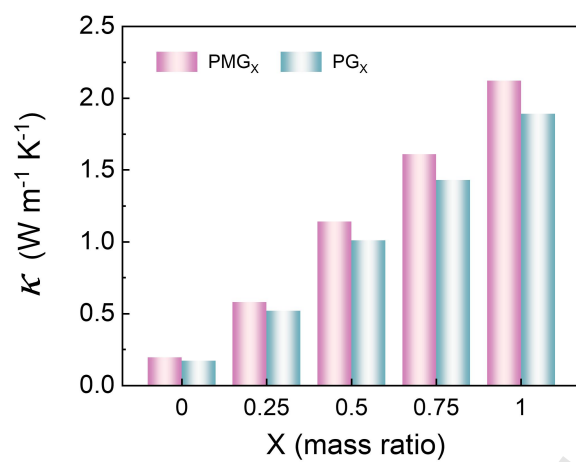


Figure S5 Comparison of the κ values between PMG_x and PG_x samples.

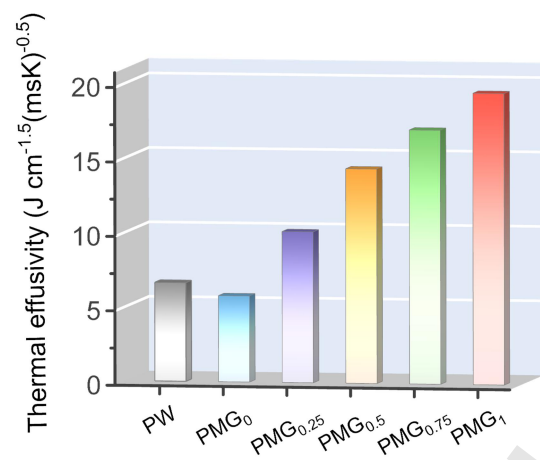


Figure S6 Thermal effusivity of PW and PMG_x samples.

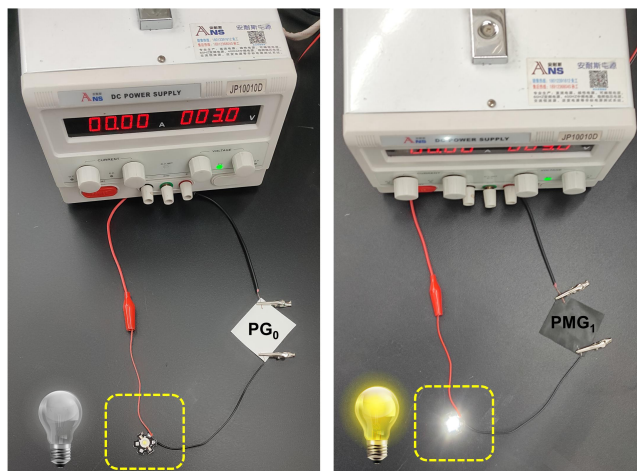


Figure S7 Comparison of the brightness of LED bulb, when using PG₀ and PMG₁ as conductors in the circuit.

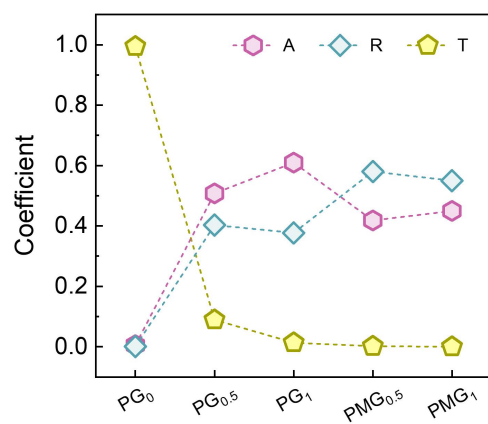


Figure S8 *A-R* coefficients of PMG_x films in X-band.

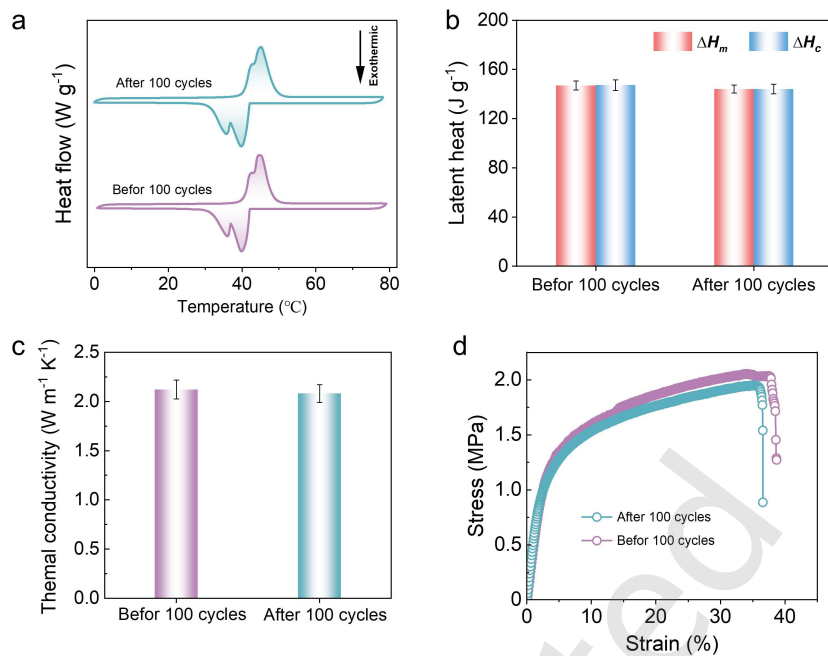


Figure S9 (a) DSC melting-solidification curves, (b) enthalpy density, (c) thermal conductivity and (d) mechanical response of PMG₁ before and after 100 thermal cycles.

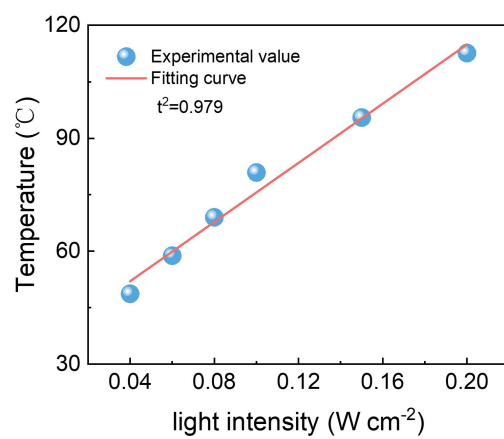


Figure S10 Experimental data and linear fitting of saturation temperature versus the irradiated light intensity.

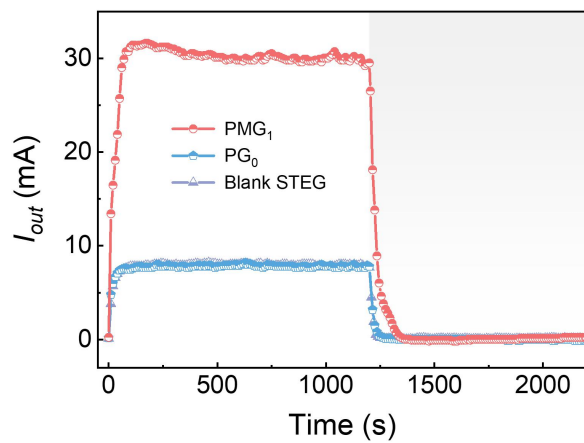


Figure S11 Output current of blank, PG₀-, and PMG₁-supported STEG under the light intensity of 0.2 W cm⁻².



Figure S12 The PMG₁-loaded STEG can energize various electronic equipment, including the LED, mobile phone, and smart bracelet with the passive voltage booster.

Supplementary Tables**Table S1** Latent heat of PW and PMG_x films.

Samples	T_c (°C)	ΔH_c (J g ⁻¹)	T_m (°C)	ΔH_m (J g ⁻¹)
PW	36.44	252.2	45.05	249.8
PMG ₀	39.15	162.7	45.73	161.3
PMG _{0.25}	39.14	160.9	45.03	158.7
PMG _{0.5}	39.84	157.4	45.04	157.1
PMG _{0.75}	40.16	152.5	44.84	151.2
PMG ₁	39.86	147.2	44.89	146.9

Here, ΔH_c / ΔH_m is solidification/fusion enthalpy. T_c / T_m is peak solidifying/melting temperature.

Table S2 Latent heat of PMG₁ film against 52th melting-solidifying cycles.

Cycle	T_c (°C)	ΔH_c (J g ⁻¹)	T_m (°C)	ΔH_m (J g ⁻¹)	Cycle	T_c (°C)	ΔH_c (J g ⁻¹)	T_m (°C)	ΔH_m (J g ⁻¹)
1	39.86	147.2	44.89	146.9	27	39.76	146.9	44.84	146
2	39.68	146.9	44.92	146.3	28	39.74	146.8	44.84	146.6
3	39.68	147.2	44.91	145.8	29	39.76	146.4	44.84	146.1
4	39.7	146.9	44.9	146.6	30	39.77	146.6	44.86	146.3
5	39.7	146.3	44.89	145.8	31	39.76	146.9	44.84	145.8
6	39.71	146.9	44.9	146	32	39.76	146.5	44.84	146.5
7	39.73	147	44.9	146.1	33	39.76	146.4	44.88	145.9
8	39.73	146.9	44.88	146.3	34	39.74	146.6	44.84	146.4
9	39.73	146.5	44.88	146.2	35	39.77	146.7	44.84	145.9
10	39.74	146.7	44.88	146.1	36	39.74	147	44.84	146.1
11	39.73	146.8	44.87	146.2	37	39.76	146.9	44.86	146
12	39.73	146.8	44.86	146.5	38	39.77	146.8	44.84	146.2
13	39.74	146.4	44.88	146.2	39	39.77	146.9	44.87	145.8
14	39.76	146.6	44.86	146	40	39.76	146.6	44.84	146
15	39.75	146.8	44.85	146.2	41	39.76	146.9	44.83	146.4
16	39.76	146.3	44.86	145.8	42	39.76	146.7	44.82	145.9
17	39.76	146.4	44.86	146.2	43	39.78	147.1	44.84	146
18	39.75	147	44.88	146.5	44	39.78	146.8	44.82	145.9
19	39.76	146.9	44.9	146	45	39.78	146.9	44.8	146.4
20	39.72	146.8	44.88	146.2	46	39.78	146.2	44.82	146
21	39.72	147	44.86	146.2	47	39.78	147	44.8	146.1
22	39.74	146.8	44.86	146.5	48	39.78	146.8	44.82	146.3
23	39.72	146.8	44.87	146	49	39.78	147	44.88	146.1
24	39.74	146.9	44.88	146.7	50	39.76	147.5	44.82	145.8
25	39.72	146.9	44.86	146.4	51	39.78	147.3	44.82	147
26	39.74	147	44.84	145.8	52	39.8	147.2	44.82	146.3

Here, $\Delta H_c/\Delta H_m$ is solidification/fusion enthalpy. T_c/T_m is peak solidifying/melting temperature.

Table S3 Comparison of our PMG₁ film with the latest reported PCCs in terms of the κ , SE_T/t , ΔH , A and σ .

PCCs	κ^a (W m ⁻¹ K ⁻¹)	SE_T/t^b (dB mm ⁻¹)	ΔH^c (J g ⁻¹)	A^d	σ^e (S m ⁻¹)	Year ^{Ref.}
MF@MA/PEG	0.19	6.1	141.3	0.691	75.3	2022 ^[4]
MPA χ -PEG	0.42	58.5	148.9	0.004	223.7	2022 ^[5]
PP/CNTs/Fe ₃ O ₄ /PW	0.59	20.6	133.8	0.375	31.0	2022 ^[6]
MX/Ni@MS/RCG/PEG	0.47	16.3	154.3	0.250	329.8	2023 ^[7]
PEG@CPGxAy	0.84	11.8	167.6	0.455	55.1	2023 ^[8]
MXene@PEG	0.95	215.5	153.7	0.054	0.09	2023 ^[9]
MXene/PA/Wood/PEG	0.82	14.8	135.5	/	/	2024 ^[10]
MPC/CuS/PEG	1.07	38.8	146.8	0.166	126.4	2024 ^[11]
CuS/ZnO/SA	0.45	40.0	155.0	/	312.0	2024 ^[12]
CNT/MXene/PEG	0.64	38.2	183.4	0.480	95.5	2025 ^[13]
MXene/PVA/PEG	1.38	43.0	175.0	/	/	2025 ^[14]
CNT/EGN/PW	2.26	18.8	160.0	/	/	2025 ^[15]
MXene/PEO@PEG	0.37	30.1	176.0	/	/	2026 ^[16]
RCP/PEG	0.39	25.0	169.6	/	/	2026 ^[17]
CNF/WP/PEG	0.35	33.9	130.2	/	1.73	2026 ^[18]
PMG ₁	2.12	194.5	146.9	0.450	302.4	This work

^a κ : thermal conductivity (W m⁻¹ K⁻¹).^b SE_T/t : total EMI shielding effectiveness/thickness (dB mm⁻¹).^c ΔH : enthalpy density (J g⁻¹).^d A : absorption coefficient.^e σ : electrical conductivity (S m⁻¹).

Table S4 Comparison in terms of the output voltage of our PMG₁-loaded STEG with latest reported counterparts.

Light intensity (W cm ⁻²)	Output voltage (mV)	Year	Ref.
0.5	220.0	2021	[19]
0.2	410.0	2022	[20]
0.8	287.0	2022	[21]
0.2	206.0	2023	[22]
0.3	423.0	2023	[23]
0.4	364.0	2024	[24]
1.4	487.1	2024	[25]
0.1	197.5	2025	[26]
0.1	129.3	2025	[27]
0.3	289.0	2025	[28]
1.5	421.6	2025	[29]
0.1	288.0	2026	[30]
0.2	454.3	2026	[31]
0.3	438.5	2026	[32]
0.5	349.6	2026	[33]
0.5	382.1	2026	[34]
0.5	670.3	This work	

References

- [1] Shahzad, F.; Alhabeab, M.; Hatter, C. B.; Anasori, B.; Hong, S. M.; Koo, C. M.; Gogotsi, Y. Electromagnetic interference shielding with 2D transition metal carbides (MXenes). *Science* **2016**, *353*, 6304.
- [2] Tan, X.; Liu, T. H.; Zhou, W.; Yuan, Q.; Ying, J.; Yan, Q.; Lv, L.; Chen, L.; Wang, X.; Du, S.; Wan, Y. J.; Sun, R.; Nishimura, K.; Yu, J.; Jiang, N.; Dai, W.; Lin, C. T. Enhanced electromagnetic shielding and thermal conductive properties of polyolefin composites with a $\text{Ti}_3\text{C}_2\text{T}_x$ MXene/graphene framework connected by a hydrogen-bonded interface. *ACS Nano* **2022**, *16*, 9254–9266.
- [3] Jin, L.; Cao, W.; Wang, P.; Song, N.; Ding, P. Interconnected MXene/graphene network constructed by soft template for multi-performance improvement of polymer composites. *Nanomicro Lett.* **2022**, *14*, 133.
- [4] He, Y. J.; Shao, Y. W.; Xiao, Y. Y.; Yang, J. H.; Qi, X. D.; Wang, Y. Multifunctional phase change composites based on elastic MXene/silver nanowire sponges for excellent thermal/solar/electric energy storage, shape memory, and adjustable electromagnetic interference shielding functions. *ACS Appl Mater Interfaces*. **2022**, *14*, 6057-6070.
- [5] Xiao, Y.-Y.; He, Y.-J.; Wang, R.-Q.; Lei, Y.-Z.; Yang, J.-H.; Qi, X.-D.; Wang, Y. Mussel-inspired strategy to construct 3d silver nanoparticle network in flexible phase change composites with excellent thermal energy management and electromagnetic interference shielding capabilities. *Compos Part B-Eng.* **2022**, *239*, 109962.
- [6] Li, X.; Sheng, M.; Gong, S.; Wu, H.; Chen, X.; Lu, X.; Qu, J. Flexible and multifunctional phase change composites featuring high-efficiency electromagnetic interference shielding and thermal management for use in electronic devices. *Chem Eng J.* **2022**, *430*, 132928.
- [7] Cheng, Z.; Chang, G.; Xue, B.; Xie, L.; Zheng, Q. Hierarchical Ni-plated melamine sponge and MXene film synergistically supported phase change materials towards integrated shape stability, thermal management and electromagnetic interference shielding. *J Mater Sci Technol.* **2023**, *132*, 132-143.
- [8] Wu, C.; Zeng, L.; Chang, G.; Zhou, Y.; Yan, K.; Xie, L.; Xue, B.; Zheng, Q. Composite phase change materials embedded into cellulose/polyacrylamide/graphene nanosheets/silver nanowire hybrid aerogels simultaneously with effective thermal management and anisotropic electromagnetic interference shielding. *Adv Compos Hybrid Mater.* **2023**, *6*, 31.
- [9] Xu, D.; Huang, Q.; Yang, L.; Chen, Y.; Lu, Z.; Liu, H.; Han, P.; Guo, L.; Wang, C.; Liu, C. Experimental design of composite films with thermal management and electromagnetic shielding properties based on polyethylene glycol and MXene. *Carbon* **2023**, *202*, 1-12.
- [10] Chen, Y.; Meng, Y.; Zhang, J.; Xie, Y.; Guo, H.; He, M.; Shi, X.; Mei, Y.; Sheng, X.; Xie, D. Leakage proof, flame-retardant, and electromagnetic shield wood morphology genetic composite phase change materials for solar thermal energy harvesting. *Nano-Micro Letters* **2024**, *16*, 196.
- [11] Wang, S.; Sun, Z.; Wang, Y.; Liang, T.; Wang, B.; Fan, C.; Chen, Y.; Liu, C.; Huang, Q. Design of CuS composite carbon-based Ni Al-IDH multifunctional phase change composite with electromagnetic shielding performance and heat storage capacity. *Chem Eng J.* **2024**, *491*, 151960.
- [12] Zhou, Z.; Huang, Y.; Shen, Q.; Li, Y.; Cheng, X. Composite phase change materials with carbon-mesh/CuS/ZnO interface biocarbon skeleton for solar energy storage, solar photocatalysis and electromagnetic shielding. *J Storage Mater.* **2024**, *90*, 111937.
- [13] Yang, J.; Wang, S.; Zhao, J.; Wang, S.; Wang, S.; Peng, Y.; Huang, Q.; Liu, C. Preparation of MXene-based aerogel-derived phase change materials for mitigating thermal runaway risk in batteries. *Chem Eng J.* **2025**, *511*, 162039.
- [14] Liu, C.; Bai, X.; Cao, Q.; Wang, S.; Huang, Q. Safety properties of 3D framework loaded polyethylene glycol composite phase change materials with MXene. *Journal of Safety and Sustainability* **2025**, *2*, 171-180.
- [15] Hu, X.; Ai, B.; Quan, B.; Shi, Z.; Liu, Z.; Sheng, M.; Ding, Y.; Yan, H.; Xiao, Y.; Lu, G.; Liu, S.; Wu, H.; Huang, X.; Yan, X.; Lu, X.; Qu, J. Synergistically networked 1D/2D/3D phase change composites with ultra-fast contactless thermal energy conversion and enhanced electromagnetic interference shielding performance. *Chem Eng J.* **2025**, *504*, 158915.
- [16] Bai, X.; Wang, S.; Guo, J.; Liu, Z.; Huang, Q.; Chen, Q.; Xie, J.; Ding, H.; Zhou, T.; Liu, C. Multifunctional polymer composite phase change material with efficient thermal management and electromagnetic shielding properties. *Polymer* **2026**, *345*, 129603.
- [17] Hu, J.; Zou, M.; Wang, J.; Zou, S.; Liu, Y.; Ma, Y.; Chen, W.; Li, Q.; He, Y.; Jiang, X.; Hu, X. Multifunctional anisotropic PEG/P-N co-doped rGO aerogel phase change composites: Tuning phonon transport for smart thermal and emi management. *Chem Eng J.* **2026**, *535*, 175841.
- [18] Yang, L.; Wang, Y.; Chen, M.; Li, D.; Wang, F.; Zhang, L.; Liang, Y.; Chen, R.; Li, J. Waste paper-spent coffee grounds synergistically derived carbonized cellulose aerogel for multifunctional phase change materials: Enhancing thermal energy storage, joule/solar heating, and electromagnetic shielding. *Biomass Bioenergy* **2026**, *212*, 109272.
- [19] Shen, Z.; Kwon, S.; Lee, H. L.; Toivakka, M.; Oh, K. Cellulose nanofibril/carbon nanotube composite foam-stabilized paraffin phase change material for thermal energy storage and conversion. *Carbohydr Polym.* **2021**, *273*, 118585.
- [20] Zhou, Y.-C.; Yang, J.; Bai, L.; Bao, R.-Y.; Yang, M.-B.; Yang, W. Super-flexible phase change materials with a dual-supporting effect for solar thermoelectric conversion in the ocean environment. *J Mater Chem A.* **2022**, *11*, 341–351.
- [21] Tang, L. S.; Zhou, Y. C.; Zhou, L.; Yang, J.; Bai, L.; Bao, R. Y.; Liu, Z. Y.; Yang, M. B.; Yang, W. Double-layered and shape-stabilized phase change materials with enhanced thermal conduction and reversible thermochromism for solar thermoelectric power generation. *Chem Eng J.* **2022**, *430*, 132773.
- [22] Zhu, C.; Hao, Y.; Wu, H.; Chen, M.; Quan, B.; Liu, S.; Hu, X.; Liu, S.; Ji, Q.; Lu, X.; Qu, J. Self-assembly of binderless MXene aerogel for multiple-scenario and responsive phase change composites with ultrahigh thermal energy storage density and exceptional electromagnetic interference shielding. *Nano-Micro Letters.* **2023**, *16*, 57.
- [23] Zhang, D.; Zhang, S.; Liang, Q.; Song, J.; Guan, M.; Zhang, T.; Chen, S.; Wang, H. One - step synthesis of multifunctional bacterial cellulose film - based phase change materials with cross - linked network structure for solar–thermal energy conversion, storage, and utilization. *Small.* **2023**, *20*, 2307259.
- [24] Hu, B.; Guo, H.; Cui, Y.; Li, J.; Cao, M.; Qi, W.; Cao, X.; Li, B. Engineering multifunctional phase change composites enabled by dual-interpenetrating hybrid scaffold for excellent thermal conductivity and electromagnetic absorption. *Chem Eng J.* **2024**, *492*, 152259.
- [25] Hu, B.; Guo, H.; Li, T.; Li, J.; Cao, M.; Qi, W.; Li, X.; Li, B. Millefeuille-like phase change composite with multi-level oriented architecture for efficient thermal management and electromagnetic interference shielding. *Chem Eng J.* **2024**, *483*, 149240.

- [26] Chen, K.; Li, Y.; Feng, Y.; Li, X.; Li, Z.; Liu, S.; Ge, C.; Chen, X. In situ vertical alignment of MoS₂ on Co/C dodecahedron boosting phase change materials for solar-thermoelectric generation. *Journal of Energy Chemistry* **2025**, *107*, 548-557.
- [27] Huang, Z.-J.; Liu, Y.-L.; Yang, J.-H.; Sun, D.-X.; Qi, X.-D.; Wang, Y. Pine needle-like hierarchical copper foam-based high-performance phase change composites for all-weather solar-thermal-electric conversion. *Chem Eng J.* **2025**, *514*, 163268.
- [28] Fang, Y.; Huang, X.; Zhao, X.; Hu, X.; Lu, G.; Shi, Q.; Ren, J.; Qu, J.; Lu, X. Renewable cellulose-reinforced flexible Ti₃C₂T_x-based phase change composites for electromagnetic shielding and thermal management. *Chem Eng J.* **2025**, *525*, 170655.
- [29] Hu, B.; Guo, H.; Li, T.; Cao, X.; Cao, M.; Qi, W.; Cui, Y.; Li, B. Engineering tiramisu-like phase change nanocomposite for superior thermal energy management and electromagnetic interference shielding. *J Mater Sci Technol.* **2025**, *206*, 113-124.
- [30] Sheng, M.; Rong, H.; Zhou, C.; Sheng, X.; Hu, X.; Tang, X.; Qu, J.; Lu, X. Simultaneously achieving high energy density, mechanical robustness, and closed-loop recyclability in phase change materials for advanced thermal energy systems. *Adv Energy Mater.* **2026**, e70874.
- [31] Shu, C.; Zhao, H.-Y.; Lu, X.-H.; Sun, X.; Liu, J.; Li, X.; Yu, Z.-Z. APP/GNP aerogel-enhanced layered phase change composites for multi-scenario energy conversion and storage. *Chem Eng J.* **2026**, *534*, 174841.
- [32] Feng, Y.; Zhao, J.; Chen, K.; Li, Y.; Liu, J.; Yu, H.; Qin, M.; Jia, K.; Han, H.; Shen, Z.; Ma, M.; Chen, X. CuS-bridged MXene-based photoresponsive phase change materials enabling thermoelectric cogeneration and microwave absorption. *Nano-Micro Letters* **2026**, *18*, 428.
- [33] Yuan, S.; Zhu, J.; Li, Y.; Li, P.; Feng, Y.; Shen, Z.; Wang, Z.; Li, R.; Yao, K.; Wu, J.; Hai, G.; Chen, X. Modulating photons and phonons in graphene - based phase change materials for solar - thermal - electrical generation. *Small* **2026**, e73606.
- [34] Li, Y.; Li, P.; Yuan, S.; Feng, Y.; Chen, X. Two birds with one stone: Integrating thermoelectric and antibacterial in bioinspired photoresponsive phase change trigger. *Nano Energy* **2026**, *155*, 112102.