Controllable Fe ion-anchored Graphene Heterostructures for Robust and Highly Thermal Conductive Cellulose Nanofiber Composites

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Research Article

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Abstract

Developing the polymer-based thermal interface materials (TIMs) is one of the most promising approaches to address heat accumulation along with the functionalization, integration, and miniaturization of modern electronics, while it is still a great challenge to balance the thermal conductivity and mechanical properties. In this article, Fe ion-anchored graphene (FeG) is successfully fabricated by a facile in situ Fe reduction of graphene oxide (GO) approach, and then cellulose nanofiber (CNF)/FeG composites are prepared by vacuum-assisted filtration. FeG exhibits excellent dispersion and exfoliation in CNF/FeG composites, due to the strong interfacial interaction between CNF and FeG, such as hydrogen bonds and “Fe-O” complex binding. Thus, CNF/FeG composite has the largely improved thermal conductivity up to 30.2 W/mK at FeG content of 50 wt%, which is substantially increased by 1160% in comparison with that of pure CNF. In addition, the mechanical performances of CNF/FeG-50 are unexpectedly simultaneously enhanced to 244 MPa for tensile strength, 4.10% for elongation at break, and 9.5 GPa for Young's modulus, outperforming pure CNF with increase of 137%, 33%, and 121%, respectively. This study provides a significant strategy for the design and construction of high thermal conductivity and high-performance polymeric TIMs in flexible and portable electronics.

Introduction

Heat accumulation causes serious efficiency and safety issues in rapidly growing fields, such as aerospace, 5G communication, electric vehicles, and sophisticated equipment manufacturing (Song et al. 2018; Ren et al. 2020; Chen et al. 2017). Central to the vital technical evolution is the development of high thermal conductive composites as thermal management materials (TMMs). To realize the high-thermal conductivity (K > 10 W/mK) for composite materials, the design of 3D interconnection thermal conduction paths is regarded as the key factors, which is in favor of optimum utilization of finite filler. Han et al. reported a boron nitrides (BNNS)/epoxy composite based on nacre-mimetic 3D networks, which exhibits the high thermal conductivity of 6.07 W/mK at 15 vol% BNNS loading, together with outstanding electrical resistivity and thermal stability (Han et al. 2019). Graphene, a superb thermally conducting filler, has been attracting the great interest of many researchers, due to its excellent thermal conductivity (5300 W/mK), high specific surface area and high aspect ratio (Song et al. 2018; Balandin et al. 2008; Duan et al. 2020). Wu et al. revealed a significant synergistic effect between the aligned graphene nanosheets (GNs) and 3D interconnected graphene foam (GF), which plays a key role in the formation of thermal percolation networks, leading to the thermal conductivity of 11.16 W/mK at 10.27 vol% (Wu et al. 2019).

Apart from constructing artificial thermal conduction path through ice-templated self-assembly (Wu 2019; Schiffres 2013), liquid crystal induction (Meng et al. 2018), solvothermal synthesis (Xu 2010), melt migration limitation in block materials (Wu et al. 2017), etc., adding large amounts of well-dispersed fillers in the nanocomposites without excessive total mass or special processing methods is prone to multifunctional organic-inorganic hybrids (Xu et al. 2018). In most cases, the threshold is commonly about 30 vol.% to realize the connection heat conduction path in composites (Song et al. 2018; Duan et al. 2020; Hu et al. 2017). Unfortunately, amounts of filler may be detrimental to the mechanical
performances and processibility, and lead to a large equivalent thermal resistance caused by agglomeration (Schiffres et al. 2013; Xu et al. 2018; Yao et al. 2018; Guo et al. 2019; Ma et al. 2019).

Noticeably, previous studies has found that 1D cellulose nanofiber (CNF) possesses some ability to disperse graphene in water solution and finally be combined together into a strong composite by hydrogen bonding (Jiang et al. 2018), thus eliminating the complex chemical modification process and increasing the process efficiency (Zhao et al. 2020; Zuo 2018; Malho 2012; Hamedi 2014; Xu 2018). Malho et al. first mixed CNF with multi-layer graphene, and then prepared a high-strength composite via intensely ultrasonic stripping (Malho 2012). The results demonstrated a simultaneous increase in Young’s modulus (16.9 GPa), tensile strength (351 MPa), and toughness (22.3 MJ m\(^{-3}\)) for the obtained composite, due to the unique hydrogen bonding, dipole effect, hydrophobic–hydrophobic interaction, and CH–π interaction between CNF and graphene (Hamedi 2014; Xu 2018; Chen 2018). However, with the increasing filler content, the mechanical properties can no longer keep up with high-strength applications, which is the common phenomenon in composites that lacks of specialized improvement (Malho 2012; Hamedi 2014; Xu 2018; Chen 2018). For example, Yang et al. took advantage of the synergistic performance of the 2D expandable graphite nanoplatelet (EG), 1D CNF, and flexible PEO. When the mass ratio of EG, CNF, and PEO reached to 95:5:3, the graphene-based composite displayed a significantly increased thermal conductivity and a decreased tensile strength about 63.3 MPa (Yang et al. 2019). Cui et al. successfully prepared hybrids based on nanodiamonds (NDs), graphene sheets (GSs) and CNF. The in-plane thermal conductivity of the hybrid at a filler content of 10 wt % could reach to 14.35 W/mK, and the moderate tensile strength was 90 Mpa (Cui et al. 2020). By contrast, balancing the thermal conductivity and mechanical properties under high loads is still a great challenge.

It is well known that the graphene oxides (GO) contains the ability to disperse into high-concentration water solution, owing to the electrostatic repulsion generated by the ionization of oxygen-containing functional groups that are introduced by edge oxidation of the graphene (Yuan 2013; Tian 2016; Dong 2018). Moreover, heat transport in graphene have been considered to rely on electron and phonon, therefore, the in-plane integrity lattice is regarded as the key factor to reduce phonon scattering and enhance the thermal conductivity. We propose that a non-covalent modification method focused on edge binding of reduced GO (rGO) is a feasible strategy to balance the thermal conductivity and mechanical properties. Wei et al. reported that the sp\(^2\) hybrid orbital lattice could be restored by the oxidation–reduction reaction with the participation of Fe ions, but the oxygen-containing functional group at the edge could be well remained (Fan 2011). This controlled reduction of GO has achieved satisfactory results in terms of energy storage and adsorption separation. However, the role of edge-functionalized non-covalent modification graphene in thermal conductive composites has not been detailedly explored yet.

In this paper, Fe ion-anchored graphene (FeG) is successfully in situ constructed via the controllable Fe-reduction of GO. FeG exhibits the high compatibility with CNF, owing to the intense interaction between FeG and CNF (e.g., hydrogen bonds and “Fe-O” complex bonds). Under vacuum filtration, CNF/FeG composite possess the dense and compact structures with high in-plane alignment. As a consequence,
the thermal conductivity of CNF/FeG-50 is greatly enhanced, due to the continuous thermal conduction path built by aligned FeG sheets. The mechanical performances of CNF/FeG-50 are also a lot higher than those of pure CNF, due to the superb reinforcing effect of FeG. The flexible CNF/FeG composites with both favorable thermal conductivity and mechanical properties open a potential application in microelectronics and advanced energy.

Material And Methods

Materials

The cellulose nanofiber (CNF) dispersion with a carboxyl content of approximately 1.08 mol g$^{-1}$ was provided by CANFOR Co., Ltd. (Canada). Graphene oxides (GOs) and graphene nano-plates (GNPs) were supplied by ENN Graphene Technology Co., Ltd (China). The anhydrous ethanol, Fe powder and HCl were purchased from Tianjin Kemio Chemical Reagent Co., Ltd (China). The reagents were all used without further purification.

Preparation of Fe ion-anchored graphene (FeG)

Figure 1 schematically shows the preparation method for FeG (Fan 2011). First, 10 g of graphene oxide was weighed in a wide-mouth beaker and dispersed in 1 L of distilled water using a 600 W sonication. Secondly, 50 g of 300 mesh iron powder and 30 mL of 35 wt.% HCl solution were added into the above dispersion to allow the redox reaction. After 24 h reaction, the product was washed several times with deionized water to neutralization. Lastly, FeG was obtained by freeze-drying for 48 h.

Preparation of CNF/FeG composites

The homogeneous mixture suspensions with different FeG contents (0, 10, 30, and 50 wt.%) were prepared via mechanically stirring for 30 min and subsequent ultrasonic treatment for 20 min. Afterwards, the above suspension was vacuum-filtered via a hydrophilic ester filter membrane with a pore size of 0.22 µm and diameter of 47 mm. The CNF/FeG composites and pure CNF were taken off from the substrate membranes after air-drying overnight at room temperature.

Characterizations

Scanning electron microscope (SEM) images were obtained using a JSM-7500F microscope (JEOL, Japan) at an operating voltage of 5 kV. The X-ray diffraction (XRD) patterns were recorded on an X-ray diffractometer (D/max 2200, Rigaku, Japan) in Cu Kα radiation (k = 1.5406 Å) at 40 kV, scanning from 5° to 40° at a speed of 4°/min. After drying at 50°C for 6 h, the chemical structures of hybrids were characterized through attenuated total reflectance Fourier-transform infrared (ATR-FTIR) spectroscopy (Nicolet iS50, U.S.A.) over the range of 4000–400 cm$^{-1}$ and Raman spectroscopy (Renishaw Invia, U.K.) using a Ne-Ha laser. The “in-plane” mode thermal diffusion coefficient k was measured by the laser flash Netzsch LFA 467 at 25°C. The thermal conductivity (K) was then determined as K = ρ·k·CP, where ρ is the mass density measured by drainage method and CP is the specific heat obtained by differential scanning
calorimetry (DSC). The tensile properties of the obtained composites were measured using an Instron tester (INSTRON 5865) at a loading speed of 0.5 mm/min.

**Discussion**

**Preparation and properties of FeG**

The preparation process for FeG is intuitively illustrated in Fig. 1a. During the reduction of GO powder to FeG at the presence of Fe/Fe$^{2+}$, the volume is largely expanded, which greatly enhances the specific surface area of FeG and interfacial interaction between CNF and FeG. The obtained FeG has good dispersion in CNF aqueous solution, as shown in Fig. 1c. A dense and continuous FeG coating is formed on the glass rod (Fig. 1d), which proves that the obtained FeG also has good self-assembly activity (Zhao 2017; Ramanathan 2008).

The Raman spectra of Go, FeG and GNP are comparatively shown in Fig. 2a. The peak intensity ratio $I_D/I_G$ of FeG is remarkably decreased from 1.14 to 0.86 in contrast with that of GO, which indicates an improved in-plane $\pi-\pi$ conjugated structure (Ferrari and Robertson 2000). Considering that the reduction process may lead to size reduction, the actual $I_D/I_G$ value may be even lower (Ferrari 2000). D- and G-bands of FeG shift to lower wavenumber after reduction, mainly due to the Fe–O charge transfer (Geng 2013). This implies that parts of the oxygen-containing functional groups preserved after the 24 h reduction are mainly centered on the edge of graphene (Jha 2017). Moreover, the symmetrical second-order phonon scattering peak (i.e., 2D peak) of FeG located at 2700 cm$^{-1}$ (532 nm) can be decomposed into four peaks, which indicates that a certain amount of bilayer graphene exists among FeG (Ni 2008). The XRD diagrams (Fig. 2b) show that the strong diffraction peak traced at 26° for GNP is corresponding to the plane lattice (002). However, the (002) plane diffraction peak of FeG is observed over a wide range of 21–26°, signifying a large amount of exfoliated graphene nanosheets (Cui 2011). Such a phenomenon takes place because Fe/Fe$^{2+}$ ions are largely intercalated between GO layers as the reaction proceeds (Geng 2013). From the Scherrer formula, the sharp shoulder peak at 21° in FeG XRD curve is corresponding to the planar spacing of about 4.2 Å, which indicates the existence of 1–2 layer graphene after the reduction. The SEM image of FeG is displayed in Fig. 2c. FeG exhibits a thin layer structure with many wrinkles, which indicates the pretty exfoliation of FeG (Stankovich 2006; Park 2014).

FTIR spectra are used to determine the types of oxygen-containing functional groups on FeG (Fig. 3). In GO, the broad and intense hydroxyl peak is situated at 3304 cm$^{-1}$, and the asymmetrical and symmetrical stretching vibrations of -CH$_2$ are located at 2925 and 2856 cm$^{-1}$, respectively. The prominent peak at about 1750 cm$^{-1}$ is corresponding to ester C = O stretching vibration. The characteristic peaks indexed to carboxy/carbonyl and aromatic C = C stretching vibrations are observed at 1637 and 1539 cm$^{-1}$, respectively. The peak existing at 1222 cm$^{-1}$ is assigned to epoxy C-O stretching vibration. The bands at 1150 and 1050 cm$^{-1}$ are the respective stretching vibrations of the tertiary and primary alcohol C–O (Lu 2008).
After reduction by Fe/Fe\(^{2+}\), the hydroxyl band centered at the higher wavenumber (3440 cm\(^{-1}\)) for FeG presents the observably decreased intensity, which illustrates the effective removal of oxygen-containing functional groups and the strong “Metal–Oxygen” (Fe-O) interaction (Scott 2013). Accordingly, the ester C = O and epoxy C-O bands are remarkably weakened with the reduction proceeding, because of the reaction between the epoxy groups and Lewis alkali metal ions (He 1998; Lerf 1998). Notwithstanding, the carboxy C = O and aromatic C = C are largely augmented, indicating the restored \(\pi\)-\(\pi\) conjugated structures. In addition, plenty of primary alcohol C-O bonds are converted to tertiary alcohol, possibly owing to the recovery of C-C bonds. The results infer that some certain oxygen-containing functional groups remain after Fe-reduction, most of which exists in the form of carboxyl groups at the edge of FeG sheets.

**Morphology, structures and mechanical properties of CNF/FeG composites**

To explore the morphology and microstructures of CNF/FeG composites, cross-section SEM images of CNF/FeG-50 are depicted in Fig. 4a and b. At such high FeG loading of 50 wt%, CNF/FeG-50 displays a highly aligned and ordered layer-by-layer structure, which is essential to the high performance of CNF/FeG composites. This aligned structure heavily depends on the good dispersion and exfoliation of FeG at the assistance of CNF/Fe ion. CNF with large number of hydrophilic groups (-COOH and -OH) can be easily intercalated into FeG interlamination through the induction of Fe ions, which is profitable for the pretty dispersion and stability of FeG in CNF solution (Fig. 4d). As shown in Fig. 4c, the positive Fe ions introduced on FeG also play a pivotal role in assisting FeG exfoliation via the dynamic “Metal-O” bond interaction between Fe ions and CNF in terms of CH-\(\pi\) interaction, dipole interaction, hydrophobic-hydrophobic interaction, etc. The interaction forces can act as the gripper of CNF and FeG to more effectively strip FeG sheets under shear flow.

Figure 4e and f describe the FTIR spectra of pure CNF and CNF/FeG composites. It is visually found that the characteristic -OH peak slightly blue-shifts from 3252 cm\(^{-1}\) for pure CNF to 3322 cm\(^{-1}\) for CNF/FeG-50 with the increasing FeG content (Fig. 4f), which illustrates the strong hydrogen binding between FeG and CNF. Similarly, C = O and C-O groups in carboxy moderately red-shifts with the addition of FeG content, due to a strong “Metal-Oxygen” complex bond between FeG and CNF (Fig. 4e) (Zhao 2017; Kemp 2017). In brief, the interfacial interaction between FeG and CNF in CNF/FeG composites is largely enhanced with the introduction of Fe ions, which is vital to the improvement of mechanical properties of CNF/FeG composites.

After the controllable reduction, the in-plane carbon lattice integrity of FeG is largely improved, so the stress load substantially increases in comparison with that of GO. Moreover, the complexation of Fe ions with oxygen-containing groups exerts a significant effect on the mechanical properties of CNF/FeG composites. Figure 5a-c show the tensile properties of pure CNF and CNF/FeG composites. It is obviously observed that pure CNF possesses the relatively poor mechanical poor mechanical performances, regrading strength of 103 MPa, elongation at break of 3.08%, and Young's modulus of 4.3 GPa. All the
characteristic mechanical properties of CNF/FeG composites are enhanced with the incremental FeG loadings. At a high FeG content of 50 wt%, the exceptionally improved strength, elongation at break, and Young’s modulus reach up to 244 MPa, 4.10%, and 9.5 GPa, respectively, which is substantially increased by 137%, 33%, and 121%, respectively, in contrary to those of pure CNF. This phenomenon is primely attributed to the strong interactions at CNF/FeG interfaces concerning hydrogen bond and “Metal-Oxygen” bridge binding, which has been confirmed by the analyses of Raman and FTIR spectra. Generally speaking, a handful of metal ions can be penetrated into or through the interior space of graphene layers, limited by the radius of metal ions, which results in the insufficient interfacial bridge binding (Zhang 2018; Park 2008; Gong 2017; Wan 2018; Shen and Feng 2018). However, abundant Fe ions are anchored on FeG sheets by “Fe-O” complexation bonds during the controllable Fe reduction. The EDS line scanning spectrum (Fig. 5d) unveils the high Fe element fraction of 28.4 wt% in CNF/FeG-50 composite. It is found in EDS mappings (Fig. 5e-f) that most Fe element is gathered at the FeG region, which manifests that Fe ions are firmly bonded on FeG sheet and not washed out in aqueous solution. The strongly anchored Fe ions on FeG can tightly link to CNF via “Metal-Oxygen” bonds, contributing to the reinforced mechanical properties of CNF/FeG composites.

**Analysis of thermal conductivity of CNF/FeG composites**

The in-plane thermal diffusivity and thermal conductivity of CNF/FeG are shown in Fig. 6a and c. It is visual that the in-plane thermal conductivity of pure CNF is merely 2.4 W/mK, in good agreement with previous reports (Shen 2018; Uetani 2016). The thermal conductivity of CNF/FeG is evidently enhanced with the introduction of FeG. When 10 wt% FeG is added, the thermal conductivity of CNF/FeG composite reaches up to 9.3 W/mK. However, the thermal conductivity of CNF/FeG-30 is slight changed (9.7 W/mK), compared with that of CNF/FeG-10. With the further increasing FeG content to 50 wt%, the thermal conductivity is sharply enhanced to 30.2 W/mK, showing the enhancement of 1160% in comparison with that of pure CNF, due to the complete thermal conductive paths built by the connection of FeG sheets. Considering the tiny alteration in $\rho$ and $C_P$ of CNF/FeG composites, the changing trend of thermal diffusivity $k$ is related to thermal conductivity $K$. On the basis of the LFA-467 instrument, thermal diffusivity is expressed as:

\[
k = 0.1388 \cdot \frac{L^2}{t_{1/2}}
\]

where $t_{1/2}$ is half of the time that it takes for the temperature to rise to the platform area and $L$ is the sample thickness. We gather essential experimental evidences (Fig. 6b and d) and find that CNF/FeG-50 expands the shortest $t_{1/2}$ (231 ms) to reach the temperature platform area, suggesting the highest thermal diffusivity (29.4 mm$^2$/s) and thermal conductivity (30.2 W/mK).

The heat conduction of graphene composites is enormously dependent on several key factors: (1) good exfoliation and dispersion of GE, (2) excellent oriented lamellae structures, and (3) satisfactory interface interaction (which can enhance the phonon resonance coupling). A great deal of in-plane aligned FeG with excellent exfoliation and dispersion has been identified in the developed composites. In addition,
intense interface interaction between FeG and CNF including hydrogen bond and “Fe-O” complex bridge binding is conductive to the construction of thermal conductive highway. A deductive mechanism related to the stack area of fillers is proposed in Fig. 7. The thermal conductivity of polymer composites are highly determined by the thermal conductivity of matrix, thermal conductivity of fillers, thermal resistance at matrix-filler interface, and thermal resistance at filler-filler interface. Pure CNF has good thermal conductivity of 2.4 W/mK, which is greater than that of most polymer matrix, due to the wonderful aligned structures formed during the vacuum filtration. When the filler loading is less than 10 wt%, the thermal conductivity of CNF/FeG composite is mainly leaned on the thermal conductivity and loading fraction of fillers. With the increasing FeG loading, the phonon scattering in insulating matrix is largely decreased, contributing to the improved thermal conductivity. Although the insulating bulk is reduced with the increasing FeG content at the filler range of 10–30 wt%, a large amount of FeG breaks the in-plane orientation structures of CNF. Thus, the thermal conductivity of CNF/FeG composites is slightly changed with the increasing FeG content in this region. At the FeG content higher than 30 wt%, FeG sheets begin to contact with each other and build the thermally conducting highway. In this stage, the thermal conductivity is primarily dependent on the filler-filler thermal interfacial resistance and sharply enhanced up to 30.2 W/mK.

To emphasize the advantages of CNF/FeG composites in thermal conduction and mechanical properties, we compared the developed composites with common CNF/GNP composites, which is given in Fig. 8. As above mentioned, high filler loading will cause insufficient interfacial bonding, agglomeration of fillers, and formation of air pockets, which makes the composites brittle and difficult to process. Although CNF can disperse and strip multilayer graphene to some extent (Malho 2012; Hamedi 2014; Xu 2018; Chen 2018), its effectiveness is strongly dependent on the mixing operation technology. In control experiments, CNF/GNP-50 is too brittle to fabricate the complete circle film, despite having the similar thermal conduction level (Fig. 8a and b). SEM observations are conducted to in depth investigate the difference of microstructures between CNF/FeG-50 and CNF/GNP-50 (Fig. 8c-f). The images reveal that CNF/FeG-50 is a dense and robust composite with pretty in-plane orientation structures. Nonetheless, CNF/GNO displays a coarse, loose, and disordered microstructure, giving rise to the inferior strength that cannot support tensile tests.

In order to estimate the contribution of this present investigation, the comparison of thermal conduction and mechanical properties in CNF/FeG composites with the reported values in other polymeric composites is summarized in Fig. 9. Visibly, it is quite difficult to achieve simultaneously superior thermal conductivity and mechanical properties in the case of polymer-based composites. In this effort, CNF/FeG composites display high-level thermal conductivity and mechanical properties at the same time, which surpasses most reported composites in previous literatures. The comparison results clearly illuminate that the proposed method is competitive to dramatically promote mechanical properties and thermal conductivity of CNF composites.

Conclusion
In conclusion, a creative interfacial control strategy is proposed to fabricate polymeric composites with both fascinating thermal conductivity and mechanical performance. Fe ion-anchored graphene (FeG) are produced via in situ reduction of GO with Fe/Fe$^{2+}$, followed by being incorporated into CNF matrix by a vacuum filtration method. The Raman and FTIR spectra clearly reveal that FeG is effectively reduced, and Fe ions are tightly anchored on FeG sheets via strong “Fe-O” complex bonds. FeG shows the wonderful exfoliation and dispersion with the help of CNF in aqueous solution, which is attributed to the intense interaction between FeG and CNF. Moreover, CNF/FeG composites have pretty in-plane orientation structures under the vacuum filtration. At FeG content of 50 wt%, the thermal conductivity is sharply enhanced to 30.2 W/mK, showing the enhancement of 1160% in comparison with that of pure CNF, due to the complete thermal conductive paths built by the connection of FeG sheets. In addition, CNF/FeG-50 exhibits the exceptionally simultaneous increment of tensile strength (137%), elongation at break (33%), and Young’s modulus (121%), in contrast with pure CNF. Hence, this strategy provides a valuable insight into the design and construction of high-thermal-conductivity and high-performance polymer composites for versatile application in modern electronic devices.

Declarations

Conflicts of interest We have no conflicts of interest to declare that are relevant to the content of this article. We certify that they have no affiliations with or involvement in any organization or entity with any financial interest or non-financial interest in the subject matter or materials discussed in this manuscript. We have no financial or proprietary interests in any material discussed in this article.

Human and animal rights This article does not contain any studies with human participants or animals performed by any of the authors. All the experiments were undertaken in this study comply with the current laws of the country where they were performed.

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Data availability All data generated or analyzed during this study are included in this published article.

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Figures

(a) Schematic diagram of the preparation process of FeG and CNF/FeG composites, (b) Schematic of the bridge binding between CNF and FeG, (c) Optical photo of CNF/FeG aqueous suspensions, (d) Optical photos for FeG coating on a glass rod

Figure 1
Figure 2

(a) Raman spectra and (b) XRD patterns of GO, FeG and GNP; (c) SEM image of FeG
Figure 3

FTIR spectra of FeG and GO in (a) 4000-1000 cm\(^{-1}\) and (b) 1800-1000 cm\(^{-1}\)
Figure 4

(a, b) Cross-section SEM characterizations of CNF/FeG-50 composite, (c) Schematic diagram of CNF/FeG stripping and dispersion mechanism, (d) Digital pictures of CNF and CNF/FeG aqueous dispersions, FTIR spectra of pure CNF and CNF/FeG composites in the wavenumber of (e) 1900-900 cm\(^{-1}\) and 3600-3000 cm\(^{-1}\) (f)
Figure 5

(a) photos of mechanical tensile tester and CNF/FeG composite, (b) Typical stress-strain curves of pure CNF and CNF/FeG composites, (c) The corresponding strength, elongation, and modulus, (d) the EDS line scanning spectrum of CNF/FeG-50, (e-h) EDS mapping images of CNF/FeG-50
Figure 6

(a) Thermal diffusivity and (c) thermal conductivity of CNF/FeG composites with different FeG contents, (b) The relative comparison of pulse laser response time, thermal diffusivity, and thermal conductivity for CNF/FeG composites, (d) The normalized Pulse signal-Time patterns for FeG/CNF composites with different contents
Figure 7
Schematic diagram for the thermal conduction mechanism of CNF/FeG

Figure 8
(a, b) Comparison of optical photos of CNF/FeG-50 and CNF/GNP-50 composites, (c, d) Cross-sectional SEM images of CNF/FeG-50 and CNF/GNP-50, respectively, (e, f) Partial enlarged SEM images in (c and d)
Figure 9

Comparison of thermal conductivity and tensile strength for different polymer-based composites (Luo 2017; Yao 2016; Li 2017; Chen 2020; Li 2020; Shen 2016; Yang 2017; Song 2018)

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