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# Lightweight Porous Polystyrene with High Thermal Conductivity by Constructing 3D Interconnected Network of Boron Nitride Nanosheets

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**ABSTRACT:** A composite foam consisting of foamed cross-linking polystyrene (*c*-PS) and boron nitride nanosheets (BNNSs) was synthesized, which shows a higher thermal conductivity (TC) than the corresponding solid counterparts. The BNNSs fillers are found to be aligned along the cell wall as a result of the biaxial stress field from cell expansion during the formation of 3-dimensional interconnectivity in the foams, resulting in an enhanced TC of 1.28 W/m K, near 2 and 4 times those of its solid counterpart and pure *c*-PS, respectively. It is found that the foaming-assisted formation of the filler network is an efficient strategy to improve the TC at low filler loadings in the composites. Furthermore, the composite foams exhibit low-density, rather low dielectric constants and dissipation factors at wide frequency and temperature ranges. The present work provides a novel approach to designing and preparing lightweight heat conductive polymers with low filler loadings as low-density heat management materials for potential applications in aeronautics and aerospace components.

**Keywords:** Thermal conductivity, Polymer composites, Porous structure, Boron nitride nanosheets, Dielectric properties

## 1. INTRODUCTION

The rapid development of electronics and optoelectronics such as high-power microelectronic packaging devices, LED (light emitting diode) lighting and chip encapsulation toward high speed and high-performance has demands placed on heat removal.<sup>1-3</sup> Therefore, the effective thermal management is critically important to ensure system performance and reliability, and enhance lifetime and accuracy.<sup>4-6</sup> Compared to traditional thermally conductive materials such as metal, carbon and ceramics materials, polymers have poor thermal conductivity (TC, ranging from 0.1~0.3 W/m K) because of random structures and twisting chains. However, polymers, due to their various advantages including low cost, lightweight, and easy processability, are becoming increasingly popular in heat-intensive applications such as LED housing, cell phone casing, electronic chip encapsulation, and high-power electric motor where heat accumulation can have deleterious effects.<sup>2-4</sup> These applications, along with emerging technologies such as flexible or wearable electronics, for which the requirements on mechanical flexibility and lightweight cannot be met by most conventional thermal management materials, put strong technological incentives on developing heat conductive polymers to avoid overheat in encapsulated chips or microelectronic packaging devices working under high frequencies or high voltages.<sup>7-12</sup>

Aligning the polymer chains represents an approach to achieving high TC in polymers. However, this approach requires certain fabrication techniques, such as electrospinning, nanoscale templating, mechanical stretching,<sup>13-17</sup> and the TC in these aligned polymers are only limited to the direction of chain orientation. It is quite challenging to take advantage of a single polymer chain in bulk structure to obtain high intrinsic TC polymers. Blending with high TC fillers is the most commonly used approach to enhancing TC of polymers, which can yield the values of TC ranging

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3 from 1 to 10 W/m K.<sup>18-24</sup> However, the large amount of fillers ( $\geq 65$  wt% fillers such as AlN, Si<sub>3</sub>N<sub>4</sub>,  
4  
5 Al<sub>2</sub>O<sub>3</sub>, BN or SiC) required to achieve appreciable enhancement in TC not only significantly  
6  
7 increases the material weight and cost but also leads to undesired electrical, mechanical, and optical  
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9 properties, or loss of the easy processability generally associated with polymers.<sup>20-23</sup> Therefore, it is  
10  
11 essential to develop new design and fabrication strategies of heat conductive polymer composites  
12  
13 loaded with minimum filler contents.  
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18 When the TC ratio of the filler to the polymer exceeds 1000, further increasing the TC of filler  
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20 has negligible effect on the effective TC of the composites.<sup>24</sup> Therefore, the inclusion of CNT  
21  
22 (carbon nanotube) or graphene flake with extremely high TC in polymers failed to remarkably  
23  
24 increase the TC of the polymers owing to the large thermal interfacial resistance resulting from the  
25  
26 significant phonon scattering at the interfaces<sup>[25-28]</sup> due to the unpaired phonon frequency.<sup>6, 9, 10, 12,</sup>  
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Up to now, extensive investigations on the preparation of heat conductive composites have been carried out on solid polymer matrix.<sup>1, 5, 18-32</sup> The porous polymers with low densities and reduced weights, although they are promising for applications that are sensitive to weights such as aeronautics and astronautics, have received much less attention due to their extremely low TC as a result of introduced air (i.e. 0.024 W/m K). Herein, we report the preparation and characterization of the porous polymer composites with high TC and low mass density when compared to their solid counterparts.

It was reported that foaming-induced biaxial flow could align filler particles along the cell

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3 boundary and enhance the formation of 3D network of fillers in polymer and therefore result in the  
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5 improvement in the mechanical, electrical and/or thermal properties of materials and simultaneous  
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7 suppression of the cell rupture.<sup>31, 50-52</sup> For filled polymer foams, on the one hand, TC will decrease  
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9 due to the introduction of air in matrix. On the other hand, the foaming-induced filler alignment  
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11 around expanding bubbles will enhance the formation of 3D filler networks, thereby leading to  
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13 rapid increase in TC. Therefore, the final TC depends on these two competing effects. Thus, foamed  
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15 polymers with enhanced TC could be achieved at optimal volume expansion (VE) percentage and  
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17 filler content. In this work, the boron nitride nanosheets (BNNSs)/cross-linked polystyrene (*c*-PS)  
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19 composites were prepared and foamed by using purified water as a foaming agent at a certain  
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21 temperature in confined space. The effects of filler loading, particle size and distribution, and VE  
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23 percentage on TC and dielectric properties of composites have been investigated. The obtained  
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25 structure-to-property relationships would provide guidelines to develop ultralow density polymers  
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27 with enhanced TC.<sup>52</sup>

## 2. EXPERIMENT SECTION

### 2.1. Materials

41 Preparation of BNNSs: Typically, 8 g h-BN powders (Sigma-Aldrich) were dispersed in 400 ml  
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43 Dimethylformamide (DMF) (Sigma-Aldrich) under vigorous stirring. The mixture was then  
44  
45 subjected to a 96 h tip-type sonication (175 W, 500 W×30%). The resultant mixture was first  
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47 centrifuged at 3500 rpm for 15 minutes, and the supernatant was collected. This step purified the  
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49 mixture from unexfoliated h-BN powders. Then, the supernatant was subjected to a 30-minutes  
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51 centrifugation at 10000 rpm to precipitate BNNSs. After vacuum drying overnight at 80 °C, BNNSs  
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53 were obtained.  
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### 2.2. Preparations of BNNSs/*c*-PS foam

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3 Preparation of BNNS/c-PS foam: a 20 ml slender glass bottle was loaded with 0.82 g purified  
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5 styrene (Sigma-Aldrich, 99%), 0.0082 g 2, 2'-azobis(2-methylpropionitrile) (Sigma-Aldrich, 99%),  
6  
7 0.21 g divinylbenzene (Aldrich, 80%) purified with 0.2 M NaOH (Sigma-Aldrich), 0.037 g span-80  
8  
9 (Sigma-Aldrich), and a stir bar. The contents were then mixed for about 30 minutes on a stir plate.  
10  
11 After that, different content of purified water was added to the mixture dropwisely with stirring.  
12  
13 The final mixture was stirred on a stir plate for 1 h. Next, various loading of BNNSs nanoparticles  
14  
15 was added to the solution, then the mixture was ultrasoniced for 30 min, and then stirred for 24 h.  
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17 The contents were sealed and placed into an oven at 60 °C for 24 h to finish the styrene'  
18  
19 polymerization and crosslinking reactions. The bottle was then broken to remove the foamed  
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21 sample, and it was then placed in the same oven for another 48 h until dry. The same procedures  
22  
23 were applied to fabricate both solid and foamed composites to ensure all samples experiencing the  
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25 same thermal history. For solid samples, no water was used in the mixture. The preparation  
26  
27 procedures of the composite foams is illustrated in **Figure 1**.

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36 The VE percentage of a foam is defined in Equation (1):

$$VE\% = \left( \frac{V_{foam} - V_{solid}}{V_{solid}} \right) \times 100\% \quad (1)$$

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42 where,  $V_{foam}$  is the volume of polymer foams, and  $V_{solid}$  is the volume of solid samples.

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45 In this work, the  $V_{foam}$  is closely related with the water loading, i.e., it increases with the  
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47 content of water. Thus, a desired VE value can be obtained through calculating and controlling the  
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49 water loading in the sealed bottle. Since the VE percentage of the composite foam is sensitive to the  
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51 composition of the water, so, TC of the composites can be effectively tuned by adjusting the water  
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53 concentration.

## 54 55 56 57 58 **2.3 Characterization**

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3 A scanning electron microscope (SEM, JEM-7000F, JEOL, Japan) and a field emission  
4 scanning electron microscope (FESEM, JEM-6700F, JEOL, Japan) were used to observe the  
5 microstructures in the composite samples. All samples were broken and the fractured surfaces were  
6 sputtered with a thin layer of gold to prevent charge accumulation prior to observation. A TCi Hot  
7 Disk thermal analyzer (C-Thermal, Canada) was used to measure TC of the samples using the  
8 modulated transient plane source method, which is based on a transient technique.  
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18 The temperature of the composites was recorded by an infrared thermograph (Fotric 220,  
19 China). The dielectric constant and loss were measured using an Agilent LCR meter (E4980A).  
20 Silver electrodes of a diameter of 20 mm were pasted on both sides of the samples (~3 mm) for all  
21 the electrical measurements. The temperature dependences of the dielectric constant and dissipation  
22 factor were analyzed with the frequency ranging from  $10^2$  Hz to  $10^6$  Hz and over a broad  
23 temperature range 25-180 °C with a Hewlett Packard 4284LCR meter using a 2 V bias in  
24 conjunction with a Delta Design oven model 2300 equipped with liquid nitrogen cooling system.  
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### 36 **3. Results and discussion**

#### 37 **3.1 Effects of filler loading and VE on thermal conductivity.**

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41 **Figure 2a** presents the effect of BNNSs loading on TCs of the solid and foamed polymer  
42 composites with a VE of 45% and 92%, respectively. TC increases with the filler loading ranging  
43 from 0 to 30 wt% for the solid and foamed composites. At low filler loadings, the amount of  
44 BNNSs is not sufficient to establish a heat conductive network in the matrix. Thus, TC increases  
45 rather slowly owing to enormous thermal interfacial resistance resulting from phonon scattering at  
46 the interfaces, which is detrimental to heat conduction.<sup>1</sup> With further increase of the filler content,  
47 BNNSs are able to form an interconnected network in the matrix. Consequently, the corresponding  
48 TC increases rapidly at high filler loadings, i.e., 30 wt%. At low filler loadings, TC of the foamed  
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3 polymer composites is obviously lower than that of solid counterparts because of the introduction of  
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5 air voids with ultralow TC into the matrix. With increasing filler content, the TC of foamed polymer  
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7 composites begin to approach and exceed that of solid samples. For example, at 30 wt% filler  
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9 loading, the maximum TC of the foamed polymer composites reach 1.28 W/m K, vs. 0.65 W/m K  
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11 of the solid composites with the same filler loading. A more than 97% enhancement in TC of the  
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13 foamed sample is apparently attributed to the formation of interconnected network of BNNSs in the  
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15 matrix.<sup>4-9</sup>  
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21 **Figure 2b** indicates that the VE has prominent influence on TC of the composites. The three  
22  
23 samples with various filler loadings exhibit similar behaviors in TC, i.e., TC first decreases, and  
24  
25 then increases to the maximum value with VE before it continues to decrease to a very low value.  
26  
27 For example, at 30 wt% BNNSs loading, and a VE of 45%, the highest TC of 1.28 W/m K is  
28  
29 obtained.. Our results suggest that the TC of the foamed composites mainly depends on the filler  
30  
31 loading and VE, and can exceed that of its solid counterpart at certain conditions. Thus, TC of the  
32  
33 foamed composites can be effectively tuned by controlling VE when the filler loading is greater  
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35 than the critical content.  
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41  
42 **Figure S1** (supporting information) describes the dependence of TC of the foamed composites  
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44 on the VE and filler loading. For the solid BNNSs/*c*-PS at a relatively low filler loadings, i.e., 30  
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46 wt%, the particles were randomly dispersed in the matrix, and no interconnected network of fillers  
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48 for heat flow could be established in the matrix. Phonon propagations are significantly hindered and  
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50 suppressed because of phonon scattering at the interfacial boundary between the filler and the  
51  
52 matrix, thereby leading to low TC.<sup>1, 43-47</sup> For the foamed polymer composites, apparently, due to the  
53  
54 biaxial flow of materials during the foaming process, the BNNSs probably either turn their face  
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56 fixed face orientation, or are aligned along the flow direction of materials, i.e., along the cell  
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3 boundary. Therefore, the preferential orientation of BNNS platelets along the cell wall of expanded  
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5 void cell would promote the interconnectivity of BNNSs, leading to a favorable impact on TC of  
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7 the composites owing to the established 3D pathways for phonons transfer.<sup>39, 53</sup> Compared with the  
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9 solid samples, under the same filler loading, it is the presence of foaming-induced filler alignment  
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11 that promotes the formation of 3D heat conductive network of BNNSs in *c*-PS,<sup>54</sup> which significantly  
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13 reduces the thermal interfacial resistance, and builds up the highway for phonon propagation across  
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15 the composites, also illustrated schematically in **Figure 2c**. Therefore, TCs of the foamed polymer  
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17 composites at moderate VE values are much higher than those of their solid counterparts.  
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24 As evidenced in **Figure 2(a-b)** and **Figure S1**, the foaming-induced biaxial stress field  
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26 increases with bubble expansion, thereby resulting in a higher degree of filler alignment along the  
27  
28 cell wall to establish a thermally conductive network. However, at a low VE, the filler particles are  
29  
30 too isolated to form interconnected networks for phonon propagation, thus resulting in high thermal  
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32 interfacial resistance in the composites. Moreover, the introduction of thermal insulating air (i.e.  
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34 0.024 W/m K) further significantly reduces the TC of the foamed composites compared with their  
35  
36 solid counterparts. As the VE increases, the foaming-induced alignment of BNNSs in the *c*-PS  
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38 matrix along the cell wall results in a higher probability of filler networking.<sup>55-57</sup> Once the cell  
39  
40 expansion reaches the critical level, i.e., most filler particles become preferentially aligned around  
41  
42 the expanded cell wall, the TC reaches the maximum value (as seen in **Figure 2c**). With further  
43  
44 increasing VE, excessive volume expansion would disrupt the interconnected thermally conductive  
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46 network of the fillers that are already established, resulting in a high thermal interfacial resistance in  
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48 the composites. The introduction of a large amount of thermal insulating voids into the matrix  
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50 becomes the predominant adverse factor on the TC of the foam.<sup>31</sup>  
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60 Similarly, at a moderate VE and a low filler loading, the filler particles are unable to form

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2 interconnected networks because of the insufficient amount of filler. Thus, the phonon scattering at  
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4 the interface are prominent and remarkably suppresses the phonon transfer and shorten the phonon  
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6 mean free path, resulting in depressed heat conduction.<sup>35-39</sup> Therefore, the different impacts of VE  
7  
8 on the TC of the foamed composites indicates that cell expansion-induced effect is more significant  
9  
10 in the composite foams with relatively low VE values. The mound-shape TC-VE relationship from  
11  
12 **Figure 2b** reveals that there are two or more competing factors governing the TC of the composite  
13  
14 foams. Overall, the positive effects of foaming on the TC of the foamed polymer include: 1)  
15  
16 foaming-assisted filler alignment along the cell walls, and 2) localization of BNNSs in the solid  
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18 phase of the foam *c*-PS. The negative influences of foaming on the TC includes: 1) the introduction  
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20 of thermally insulating voids in the matrix, and 2) the disruption of filler network between adjacent  
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22 cells.<sup>31, 35,47, 52</sup>  
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31 **Figure S2** (supporting information) summarizes the TCs of the composites filled with BNNSs  
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33 with different sizes (lateral size of 1  $\mu\text{m}$  and 0.5  $\mu\text{m}$ , respectively). In **Figure S2a**, the particle size  
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35 shows negligible influence on the TC of the composite foam at low filler loadings because the  
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37 population density of BNNSs is not sufficient to form interconnected networks for heat flow.  
38  
39 Therefore, the TC depends more significantly on the filler concentration. The composites with a  
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41 smaller BNNSs demonstrates more pronounced enhancement in the TC compared to those with a  
42  
43 larger filler.<sup>1</sup> This is because the smaller particles with a higher specific surface area would have a  
44  
45 higher efficiency to form robust and stable heat conductive pathways/networks at higher filler  
46  
47 loadings in the matrix.<sup>3,5</sup> As shown in **Figure S2b**, only at moderate VE values, the composites with  
48  
49 smaller particles increases more quickly with filler loading.<sup>22-26, 31</sup> However, at low and high VE  
50  
51 values, the particle size has negligible influence on the TC of the composites because there is no  
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53 network formed at these stages. Therefore, the filler size could influence the TC only when an  
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interconnected network of the particles is formed in the composites.

### 3.2 Effective filler loading

Effective filler loading in the foamed polymers can be determined by the equation:<sup>31</sup>

$$V_e = \frac{V_f(V_{solid} + V_{void})}{V_{void}} \times 100\% \quad (2)$$

where,  $V_e$  is the effective filler loading,  $V_f$  is the volume fraction of filler in the composites,  $V_{solid}$  is the volume of the solid phase in the foamed composites, and  $V_{void}$  is the total volume of all voids in the foamed composites.

According to Equation 2, the effective filler loadings for different systems are plotted in **Figure 2d**, which presents the effect of VE on the effective filler loading in solid phase. With the increase of VE, the effective filler loading exhibits a linear increase, which could result in a higher probability of filler networking, thereby enhancing the TC of the foamed composites. For instance, a 45% VE of *c*-PS foam with 15.8 vol% BNNSs results in an effective BNNSs concentration of 18.7 vol%. The localization of BNNSs in the *c*-PS matrix would result in a higher probability of filler networking, thereby improving the TC of the composite foams.

It is well known that the TC is closely related with the microstructure of foamed composites. Apart from the VE, the cell size has a prominent effect on composites' microstructure, thereby influencing the TC and other physical properties. **Figure S3** (supporting information) presents the TC dependence on cell size for the composites at various filler loadings and VE values. **Figure S3a** depicts the TC of *c*-PS with 20 wt% and 30 wt% BNNSs and controlled VE against cell size. TC is seen to first increase to a maximum value, then decrease with further increasing VE, and the maximum TC is obtained at a moderate VE for different systems. TC can be enhanced by a moderate amount of cell size growth, while, high VE, i.e., excessive cell expansion, will bring about

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2  
3 suppressed TC. For 30 wt% BNNSs/c-PS, the top TC is observed to shift towards higher VE, and  
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5 the best cell size for the three foamed composites with a VE of 22%, 45% and 92%, are 6 $\mu$ m, 8 $\mu$ m  
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7 and 18 $\mu$ m, respectively. Similiarly, the 20 wt% BNNSs/c-PS exhibit similar TC trend with cell size  
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9 distribution. At low VE value cell size has important effect on TC of composites, whereas, TC is  
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11 insensitive to the variant of cell size at high VE.<sup>1, 31</sup> As the cell size further increases, both excess  
12  
13 foam expansion and cell size growth would disrupt the formed filler interconnectivity among the  
14  
15 cell walls, thereby reducing the TC owing to the introduction of large volume of thermally  
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17 insulating. So, the different in sensitivity of TC on cell size implies that cell expansion induced  
18  
19 effects are more significant in foamed composites with high filler concentration and low VE.<sup>31, 47, 52</sup>  
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21 **Figure S3b** plots the calculated average cell size as a function of VE for c-PS with 20 wt% and 30  
22  
23 wt% of BNNSs. Cell size is seen to increase with VE for the two composites. And the maximum  
24  
25 TC is achieved at a VE of 45% for 30 wt% BNNSs/c-PS composites, the optimal cell size is about 6  
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27  $\mu$ m, which is consistent with the results in **Figure 2b** and **Figure S3b**.

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36 From Figure S3, it is found that there exists relationship between the VE and filler loading,  
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38 which causes the change to the microstructure of foamed composites, thereby inflencig TC of  
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40 composites. At low filler loading, no matter how to alter the VE value, TC changes little against  
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42 both VE and cell size. The cell size of a foam increases almost linearly with the VE, so, at large VE  
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44 or cell size, the fillers can not be connected with each, thus producing margin enhancement in TC of  
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46 samples owing to the large thermal contact resistance. Only at relatively high filler loading, the cell  
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48 size is seen to play an obvious effect on TC. So, TC is closely with the microsture of samples. with  
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50 increasing the filler loading, BNNSs begin to contact with an increase in VE and cell size, and the  
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52 network can be fromed at an optimal VE value corresponding to the best cell size, thereby  
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54 producing the maximum TC, which can not be obtained for the other canses such as the VE = 0 or  
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92%. Therefore, in order to obtain a maximum TC, the microstructure with connected BNNSs inside the cell wall should be created and developed by virtue of adjusting the VE, filler loading and cell size parameters.

### 3.3 Simulation results

The thermal transport process in the composite can be obtained by solving the heat conduction equation:

$$\frac{\partial}{\partial x_i} \left[ k_{ij}(\mathbf{x}) \frac{\partial T(\mathbf{x})}{\partial x_j} \right] + q(\mathbf{x}) = \rho(\mathbf{x}) c_p(\mathbf{x}) \frac{\partial T(\mathbf{x})}{\partial t} \quad (3)$$

where  $k_{ij}(\mathbf{x})$ ,  $\rho(\mathbf{x})$ ,  $c_p(\mathbf{x})$ , and  $T(\mathbf{x})$  represent the spatial dependent TC tensor, mass density, constant-pressure heat capacity, and temperature. For the TC, isotropic values of 0.2, 350, and 0.024 W/m K are assigned for the polymer phase, BNNSs phase, and pore phase, respectively. The densities of the polymer phase, BNNSs phase, and pore phase are assumed as 1.0 g/cm<sup>3</sup>, 2.1 g/cm<sup>3</sup>, 0.0012 g/cm<sup>3</sup>, respectively. Heat capacities of 1.3 J/(g K), 0.793 J/(g K), and 0.718 J/(g K) are used for the polymer phase, BNNSs phase, and pore phase, respectively. For the composite with a specific microstructure, the temperature distribution  $T(\mathbf{x})$  and heat flux  $J_i$  at the steady state of Equation 3 are solved using the phase-field Spectral Iterative Perturbation Method,<sup>56, 59</sup> from which the effective TC can be calculated according to  $k_{ij}^{\text{eff}} = -\langle J_i \rangle / \langle \partial T(\mathbf{x}) / \partial x_j \rangle$  with  $\langle \rangle$  representing the volumetric average.

**Figure 3** depicts the several microstructure models of BNNSs/*c*-PS with various VE and the simulation results of their effective TCs and energy flux distributions. With expanding the total volume of the samples, the volume fraction of pores increases and the volume fraction of BNNSs decreases. When the expanded volume is about 111% times of the pristine sample, BNNSs are dispersed in *c*-PS matrix. The effective TC decreases to be around 0.26 W/m K as a result of the introduced pores. With continuous increase of the total volume to 146% times of the pristine sample,

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2 more micro-pores are introduced and BNNSs are driven to be around the pore surfaces. As a result,  
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4 BNNSs are connected with each other to form a continuous shell layer during the foaming process.  
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7 The connection of the shell layers makes a continuous path for conducting thermal energy, which  
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10 can be seen from the thermal energy flux distributions. However, with expanding the volume to be,  
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12 e.g. 373% times of the pristine sample, the total volume fraction of BNNSs becomes very small (i.e.  
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14 2.68%). In this case, BNNSs are unable to form continuous thermal conductive path and only  
15  
16 randomly dispersed around the pore walls. Therefore, the effective TC of the excessively expanded  
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18 sample will be even much smaller than the pristine sample. The simulation results shown in **Figure**  
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21 **3** agree well with the experimental data as shown in **Figure 2b**.

### 22 23 24 25 26 **3.4 Heat dissipation performance**

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28 In order to demonstrate the thermal management applications of the BNNSs/*c*-PS porous  
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30 composites, the variations of surface temperature of the composites with time during cooling were  
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32 recorded by an infrared thermal imager. The samples of pure *c*-PS, 30 wt% BNNSs/*c*-PS solid  
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34 composites (VE=0), 25 wt% BNNSs/*c*-PS (VE=45%) and 30 wt% BNNSs/*c*-PS (VE=45%)  
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36 composites were employed. The variation of surface temperature with cooling time for the four  
37  
38 systems were measured from the same initial temperature. The images of temperature distribution  
39  
40 with time are shown in **Figure 4**. To investigate heat dissipation performance, all the samples were  
41  
42 placed in an oven with a temperature of 75 °C for 3 h to ensure uniform sample temperature and  
43  
44 then transferred to a thermal insulating foam stage at room temperature. As shown in **Figure 4a**,  
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46 during the heat dissipation process, the samples of 25 wt% BNNSs/*c*-PS (VE=45%), and 30 wt%  
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48 BNNSs/*c*-PS (VE=45%) composites exhibit much faster decrease with time in comparison with the  
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50 pure *c*-PS and 30 wt% BNNSs/*c*-PS solid composites containing randomly dispersed BNNSs. This  
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52 is because the composites of 25 wt% BNNSs/*c*-PS (VE=45%), and 30 wt% BNNSs/*c*-PS (VE=45%)  
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3 have much better thermal responses due to their higher TC. Taking the sample surface temperature  
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5 at 300 s as an example, they are about 19 °C and 23 °C for the 30 wt% BNNSs/*c*-PS (VE=45%), and  
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7  
8 25 wt% BNNSs/*c*-PS (VE=45%) composites, which are lower in comparison with the composites  
9  
10 of pure *c*-PS (34 °C) and solid 27 wt% BNNSs/*c*-PS (30 °C). From the infrared thermal images of  
11  
12 the four systems shown in **Figure 4b**, it is clearly observed that the results of surface temperature  
13  
14 variation are consistent of those shown in **Figure 4**. These results illustrate that the prepared  
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16 BNNSs/*c*-PS foams at moderate VE are promising for applications in novel lightweight thermal  
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18 management materials.  
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### 23 **3.5 Microstructures**

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26 **Figure 5** presents the representative SEM micrographs of the BNNSs/*c*-PS composites filled with  
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28 different filler loadings. As shown in **Figure 5a**, the filler particles were homogeneously dispersed  
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30 in the matrix of the 30 wt% BNNSs/*c*-PS solid composites (VE=0%). As a result, no heat  
31  
32 conductive pathways are formed in the matrix, which accounts for the slow enhancement in the TC  
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34 of the composites with filler loading. **Figure 5(b-d)** illustrates the microstructures of the 25 wt%  
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36 and 30 wt% BNNSs/*c*-PS composites with a VE of 45%. It is found that the BNNSs were  
37  
38 preferentially aligned around cell wall in the foamed composites and able to form a interconnected  
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40 heat conductive network for phonon transfer.<sup>32, 36</sup> Compared with the random dispersion of BNNSs  
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42 in the matrix (**Figure 5a**), at the same filler loading, VE generates biaxial stretching along their cell  
43  
44 walls, and therefore induces the preferential alignment of the platelets along the cell walls.  
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46 Consequently, the localization of BNNSs in the matrix results in a higher probability of filler  
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48 networking, and thus enhances the TC owing to the reduced phonon scattering at the filler-filler  
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50 interfaces.<sup>32, 35, 47-52</sup>  
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### 60 **3.6 Density**



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3 **Figure 6a** shows the density variations of the composites as a function of BNNSs loading and VE  
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5 percentage. With increasing filler loading, the density for all the samples increases linearly. For  
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7 example, the density of the composites increases from 1.1 to 1.2 g/cm<sup>3</sup> for the solid samples  
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9 containing 10 wt% and 27.5 wt% BNNSs, respectively. The excellent linear relationship between  
10  
11 the density and filler loading suggests that the BNNSs were homogeneously dispersed into the  
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13 aqueous suspension without precipitation during foaming process. The density of the foamed  
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15 composites is remarkably lower than that of solid samples because of the introduced air into the  
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17 composites. Furthermore, the density of the foamed composites decreases with VE percentage  
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19 under the same filler loading. The inset in **Figure 6a** displays the ultralight weight foam with 25  
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21 wt% BNNSs standing on a piece of tender leaf. For aeronautics and aerospace applications,  
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23 materials with lightweight are a prerequisite to achieve high flying speed and save energy, so, the  
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25 concurrently enhanced lightweight and TC performances of the porous BNNSs/c-PS composites'  
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27 promotes future aerospace and avionic devices' heat dissipation.  
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36 **Figure 6b** demonstrates the effect of VE on the density and the TC of the composite foam  
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38 containing 25 wt% BNNSs. It can be seen that, with the increase in VE, density reduces accordingly,  
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40 and the TC first decreases due to the introduction of air into the composites, then increases to the  
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42 maximum value before it decreases again. The maximum TC value further confirm that the biaxial  
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44 strain-induced filler alignment along cell walls is strong enough to form the interconnected  
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46 networks of filler in the foam and significantly promote the TC. Moreover, the sample with the  
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48 maximum TC possesses a lower density and a smaller VE compared to the solid sample. Therefore,  
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50 it is concluded that the lightweight polymer composites with a high TC can be obtained by using the  
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52 foaming technology.<sup>54</sup>  
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### 60 **3.7 Specific thermal conductivity**

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3 For polymer composites with potential applications in aviation and aerospace, effective weight  
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5 reduction is highly desired because the weight reduction can significantly increase the flying speed  
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7 and distance for aircraft and space rocket. In order to precisely describe the TC of foamed polymer  
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9 composites, similar to specific strength, specific modulus and specific heat capacity, a new  
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11 parameter, specific thermal conductivity (STC), i.e., TC divided by density, is suggested herein:  
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$$14 \quad STC = \kappa / \rho \quad (4)$$

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17 where,  $k$  is TC, and  $\rho$  is density.  
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21 **Figure 6(c-d)** illustrates the STC of the composites with various filler loadings and VE. From  
22  
23 **Figure 6c**, it can be found that at low filler loadings the STC of the foamed polymers is lower than  
24  
25 those of solid samples, whereas at higher filler loadings the foamed composites exhibit a much high  
26  
27 STC compared with solid samples. As shown in **Figure 6d**, the STC curves exhibit a mount-shape  
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29 against VE. At relatively low VE values, the STC increases to the maximum value, and then  
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31 reduces slowly with increasing VE. Compared with solid samples, the STC of the foamed  
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33 composites with higher filler loadings are much higher owing to their lower densities. However, at  
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35 low filler loadings, the VE has less significant influence on the STC due to the absence of 3D  
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37 network of fillers in the composites.<sup>26-30</sup> Therefore, in this study, at relatively low filler loadings, we  
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39 successfully prepared polymer composites with much high STC compared with their solid  
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41 counterparts at the same filler loading.  
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50 This kind of ultralight structures with 3D interconnected porous is propular in nature. For  
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52 example, **Figure S4** (supporting information) shows the 3D interconnected porous structures of a  
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54 loofah sponge after removal of seeds. In this work, the BNNSs/*c*-PS porous structure, silimar to the  
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56 3D interconnected loofah sponge structure, has the following advantages: 1) ultralight; 2) high  
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3 specific strength and modulus; 3) high STC, which are highly desired in the applications of  
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5 aerospace and aircraft components.  
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8 In order to demonstrate the effectiveness and superiority of the foaming-induced filler network  
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10 in enhancing the TC of the composites, the STC of this composite foams has been compared with  
11  
12 the literature results as shown in **Figure 6e**.<sup>2-4, 7-9, 12, 19-4, 26-30, 32-45,50-54</sup> Compared to the literature  
13  
14 reports on the thermally conductive polymer composites, the BNNSs/*c*-PS foam gives the best STC  
15  
16 value at low filler loadings. **Figure 6f** summarizes the enhancement of the TC (*i.e.*, the TC ratio of  
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18 composites vs. a polymer matrix) of the composites that have been reported.<sup>2-4, 7-9, 12, 19-4, 26-30,</sup>  
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Clearly, the BNNSs/*c*-PS foam exhibits the highest TC enhancement at low weight  
fractions among the reported heat conductive composites.

### 3.8 Dielectric properties

**Figure 7(a-b)** presents the relative dielectric constants ( $\epsilon$ ) and loss versus VE percentage for the 30 wt% BNNSs/*c*-PS composites across the frequency range from 200 to  $2 \times 10^6$  Hz. It can be seen that  $\epsilon$  of the composites decreases with the increase of VE. The decrease in  $\epsilon$  can be explained by the introduction of air in the composites as a result of very low  $\epsilon$  of air.<sup>35</sup> Furthermore,  $\epsilon$  for all the samples are apparently independent on frequency over the entire frequency range, suggesting that dipole polarization is the dominant mechanism and follows well with the external electric field.<sup>1</sup>  
**Figure S5** (supporting information) presents the  $\epsilon$  and loss of the 20 wt% BNNSs/*c*-PS. The  $\epsilon$  of the composites exhibits a very slight increase when the filler loading increases from 20 to 30 wt%. The loss decreases obviously due to the incorporation of air into the composites. Over the investigated frequency range, the loss is essentially lower than 0.005 for the two composites with various VE values.

**Figure 7c** presents the effective relative  $\epsilon$  versus filler loading of the BNNSs/*c*-PS composites

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3 with a VE of 92%. As the filler concentration increases, the effective relative  $\epsilon$  increases slowly,  
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5 and exhibits independent of frequency over the entire 200~2 $\times$ 10<sup>6</sup> Hz range. The effective relative  $\epsilon$   
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7 of the composites with 20 and 30 wt% BNNSs only increases from 1.6 to 1.82 at 1 kHz,  
8  
9 respectively, as compared to 2.8 and 3.1 of their solid counterparts. As seen in **Figure 7d**, the  
10  
11 BNNSs/*c*-PS composites with a VE of 92% also show very low loss over the frequency range.  
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14  
15 **Figure 8(a-b)** presents the effective relative  $\epsilon$  and loss at 1 kHz versus temperature of the  
16  
17 BNNSs/*c*-PS composites with a VE of 92%.  $\epsilon$  for pure *c*-PS and three composites increases with the  
18  
19 filler loading and rises very slowly with temperature. Only the composites with 30 wt.% BNNSs  
20  
21 has a noticeable increase in  $\epsilon$  starting at about 110 °C. As seen in **Figure 8b**, the loss also increases  
22  
23 with filler loading, and increases slowly with temperature at <110 °C. After 110 °C, a abrupt  
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25 increase of the loss to the maximum value is observed for all the composites, suggesting that the  
26  
27 temperature plays a key role in determining the dielectric loss of the composites. This transition  
28  
29 temperature actually corresponds to the glass transition temperature ( $T_g$ ) of the *c*-PS. At a  
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31 temperature near  $T_g$ , the dipoles begin to have enough mobility to contribute to the loss and  $\epsilon$ .<sup>44</sup>  
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39 **Figure 7(c-d)** shows the effective relative  $\epsilon$  and loss versus temperature of the 20 wt%  
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41 BNNSs/*c*-PS composites with various VE. With the increase of VE, all the composites exhibit an  
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43 obvious decrease in  $\epsilon$  and loss. All the  $\epsilon$  are almost independent on temperature from room  
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45 temperature to 180 °C, whereas the loss increases with temperature and exhibits a clear rise starting  
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47 from  $T_g$  owing to the enhanced mobility of polymer segments at elevated temperatures.<sup>55</sup> It can be  
48  
49 seen that compared with pure *c*-PS and the solid BNNSs/*c*-PS with the same filler loading, the  
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51 foamed composites still exhibit very low  $\epsilon$  and loss from room temperature to 180 °C as shown in  
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53 **Figure S6** (supporting information).  
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## 60 **4 Conclusion**

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3 In summary, *c*-PS foams with various BNNSs loadings at different VE percents were prepared  
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5 and characterized to investigate the effects of foam morphologies on the TC and dielectric  
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7 properties of the formed composites. It is found that the VE has a dramatical effect on the TC of the  
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9 composite foam. The optimized VE gives the composite foam with a higher TCs than their solid  
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11 counterparts. The biaxial stress field yielded during the foam formation promotes BNNSs  
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13 alignment along the cell wall and the formation of 3D interconnected fillers in the foams at the  
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15 optimized VE, thereby leading to much enhanced TC. For example, the TC of 30 wt% of  
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17 BNNSs/*c*-PS composites with a VE of 45% reach as high as 1.28 W/m K, which represents a 97%  
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19 increases over 0.65 W/m K of its solid counterpart.  
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26 In contrast, at low VE, the foams exhibit low TC owing to the absence of interconnected  
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28 network of fillers in the matrix and the introduction of thermally insulating air voids into the  
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30 composites. High VE percentage would not only result in high volume fraction of thermally  
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32 insulating air voids but also disrupt the development of continuous thermally conductive path,  
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34 leading to decreased TC. The experimental data are consistent with the simulation results.  
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39 Furthermore, the composite foams exhibit ultralow density,  $\epsilon$  and loss at wide frequency and  
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41 temperature ranges, along with ultrahigh STC. In summary, the results demonstrate that the  
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43 foaming-assisted filler networking is a feasible processing strategy to improve the TC of the  
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45 composite foams especially at low filler loadings. The polymer composite foams with low  $\epsilon$  and  
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47 loss and ultrahigh STC could provide a new material family for potential applications in aeronautics  
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49 and aerospace to address the emerging needs of thermal managements.  
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## 57 **Supporting Information**

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59 Schematic representation of the effects of VE and filler loading on the TC of foamed polymers  
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3 **(Figure S1)**; Effect of particle size on TC for foamed composites with various VE and filler  
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5 loadings **(Figure S2)**; Effect of cell size on TC of foamed composites with various VE and filler  
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7 loadings **(Figure S3)**; 3D interconnected biological porous structures found in the nature  
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9 loofah sponge **(Figure S4.)**; Dielectric properties of the 20 wt% BNNSs/*c*-PS composites as a  
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11 function of VE **(Figure S5.)**; Dielectric properties of *c*-PS and BNNSs/*c*-PS composites as a  
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13 function of VE **(Figure S6.)** (PDF)  
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### 48 **Notes**

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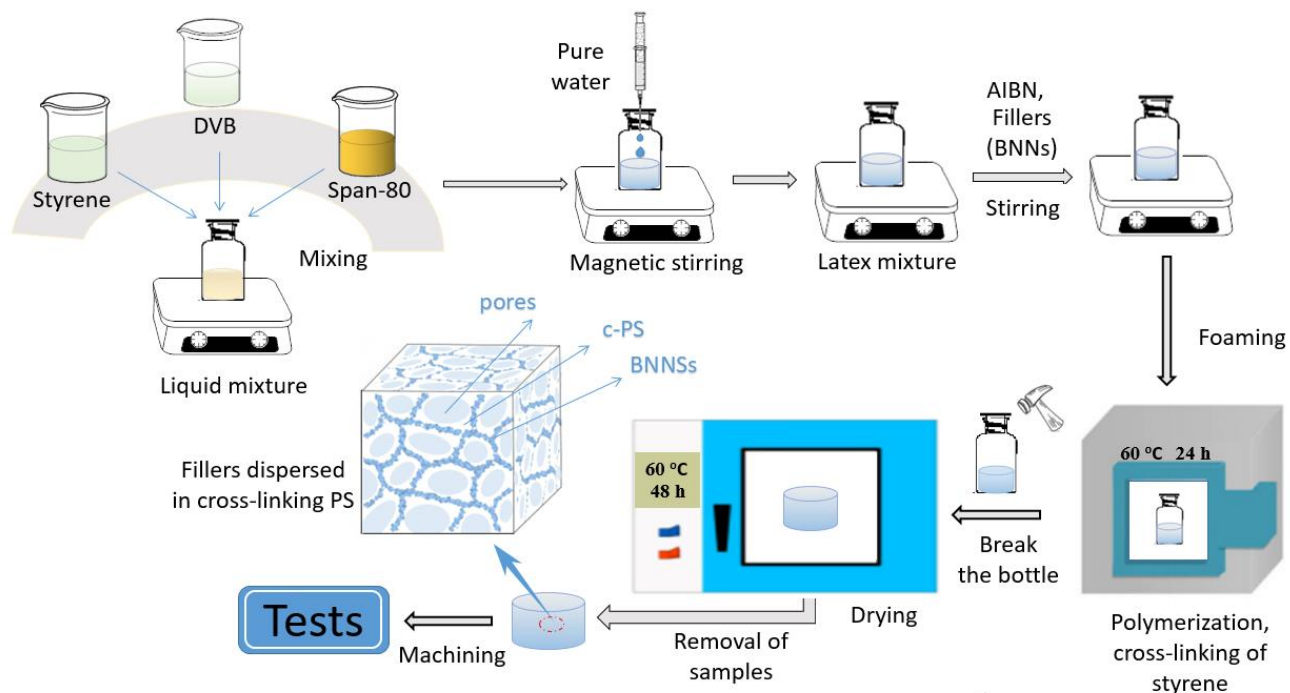
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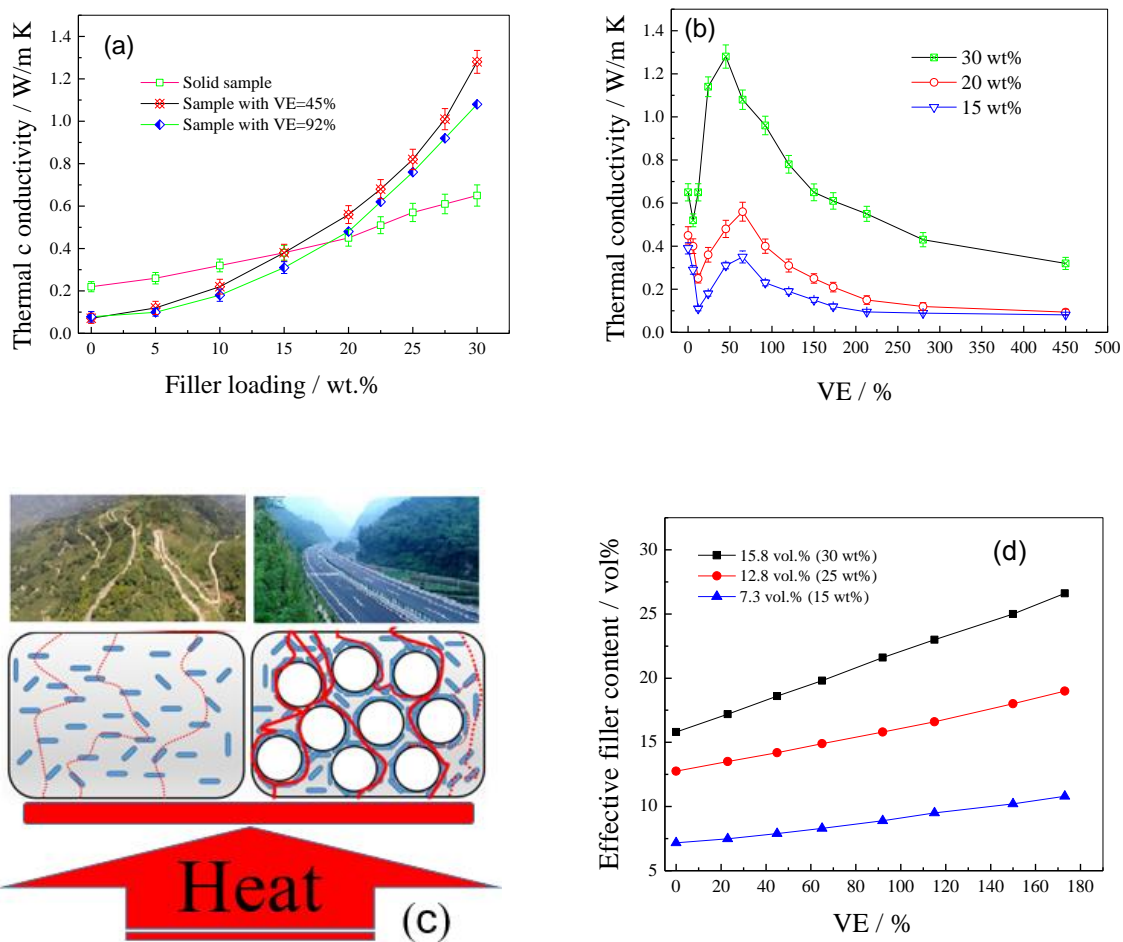
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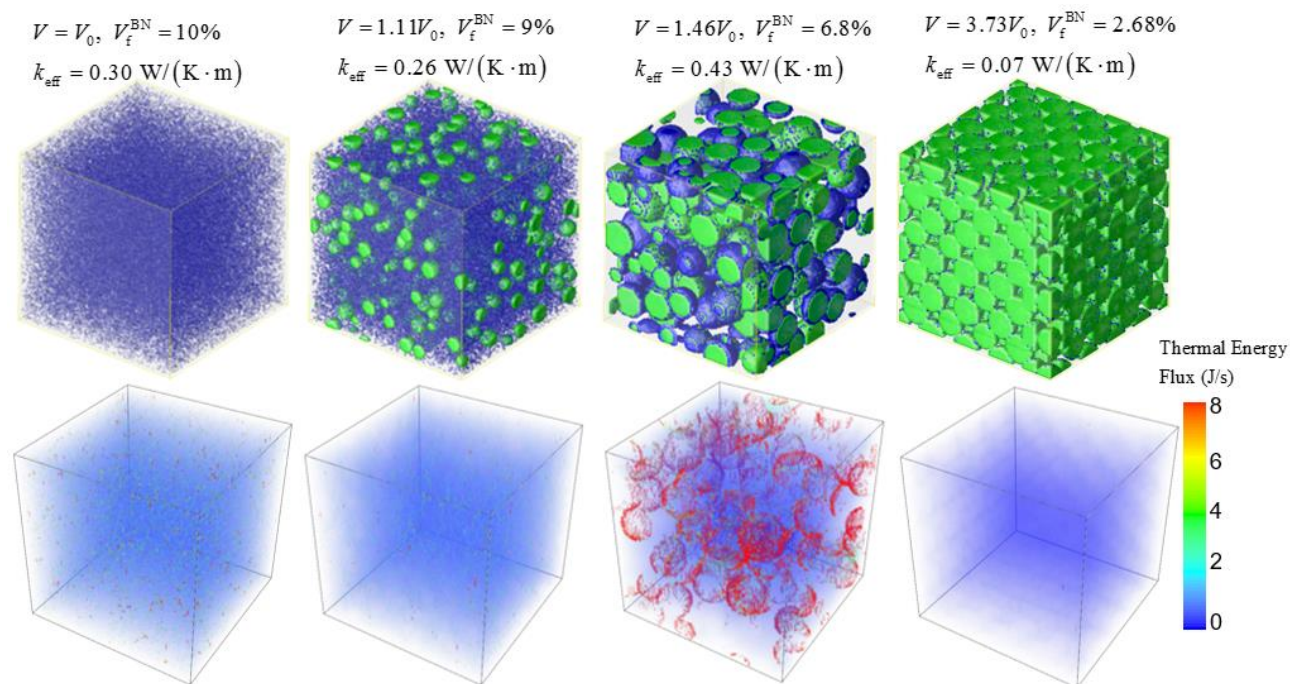


**Figure 1.** Schematic diagram of light porous cross-linking polystyrene with high thermal conductivity by constructing 3D interconnected network of boron nitride nanosheets.

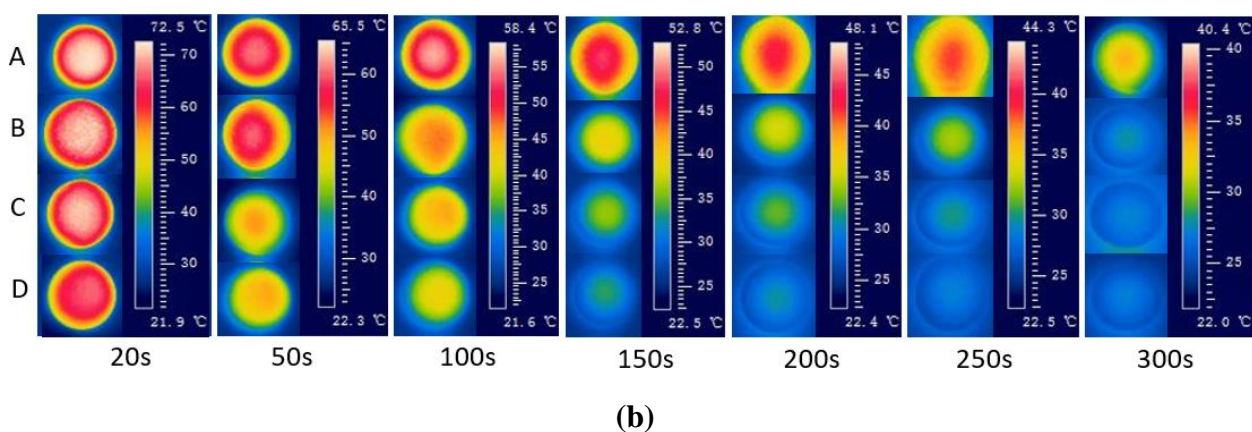
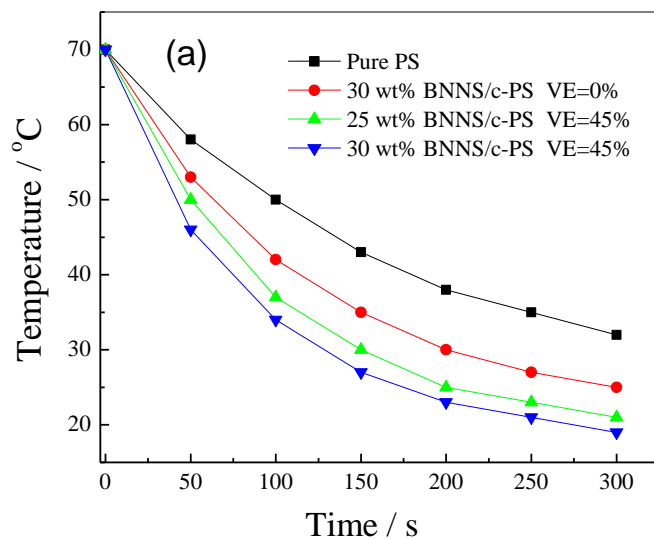


**Figure 2.** Dependence of TC of the composites on filler loading (a) and VE (b); Schematic representation of heat transfer in solid sample without heat conductive pathway, and foamed samples with heat conductive network of fillers (c); Effect of VE on effective filler loading for foamed composites (d).

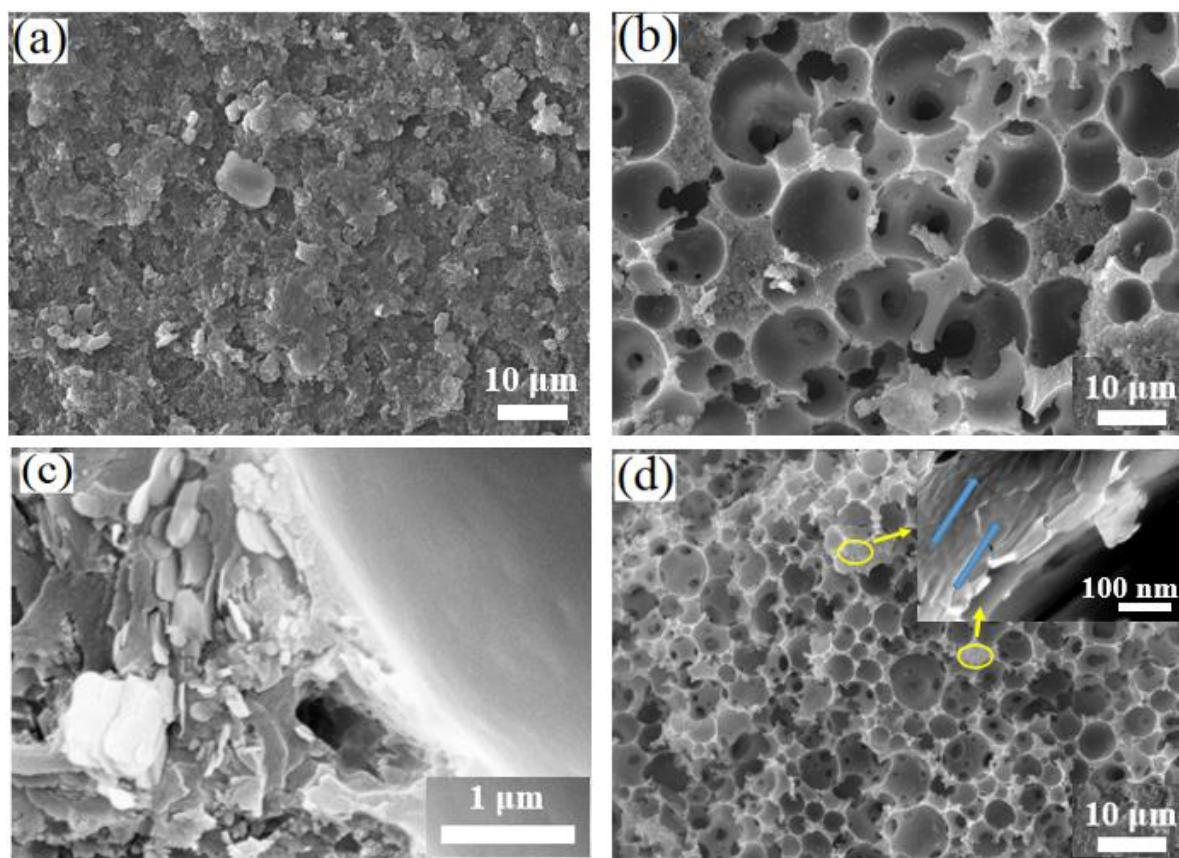




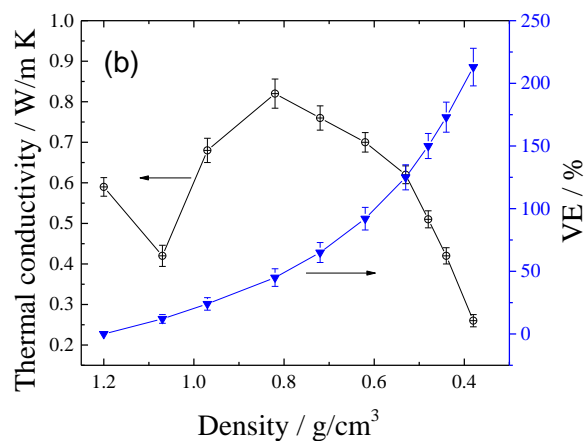
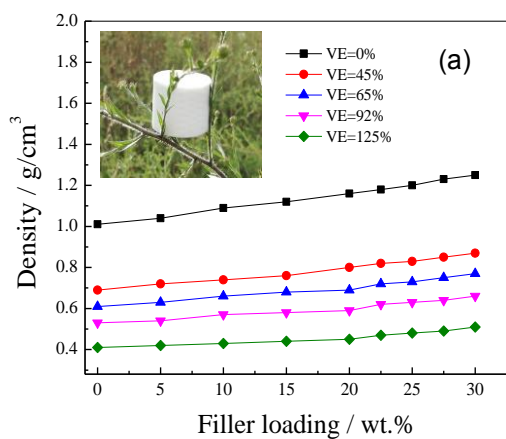
**Figure 3.** Microstructures (upper panel) and effective TC and energy flux distributions (lower panel) for the composites with different volume expansion ratios. In the microstructure, blue color and green color represent BNNSs phase and pore phase, respectively. The transparent space represents the polymer phase.

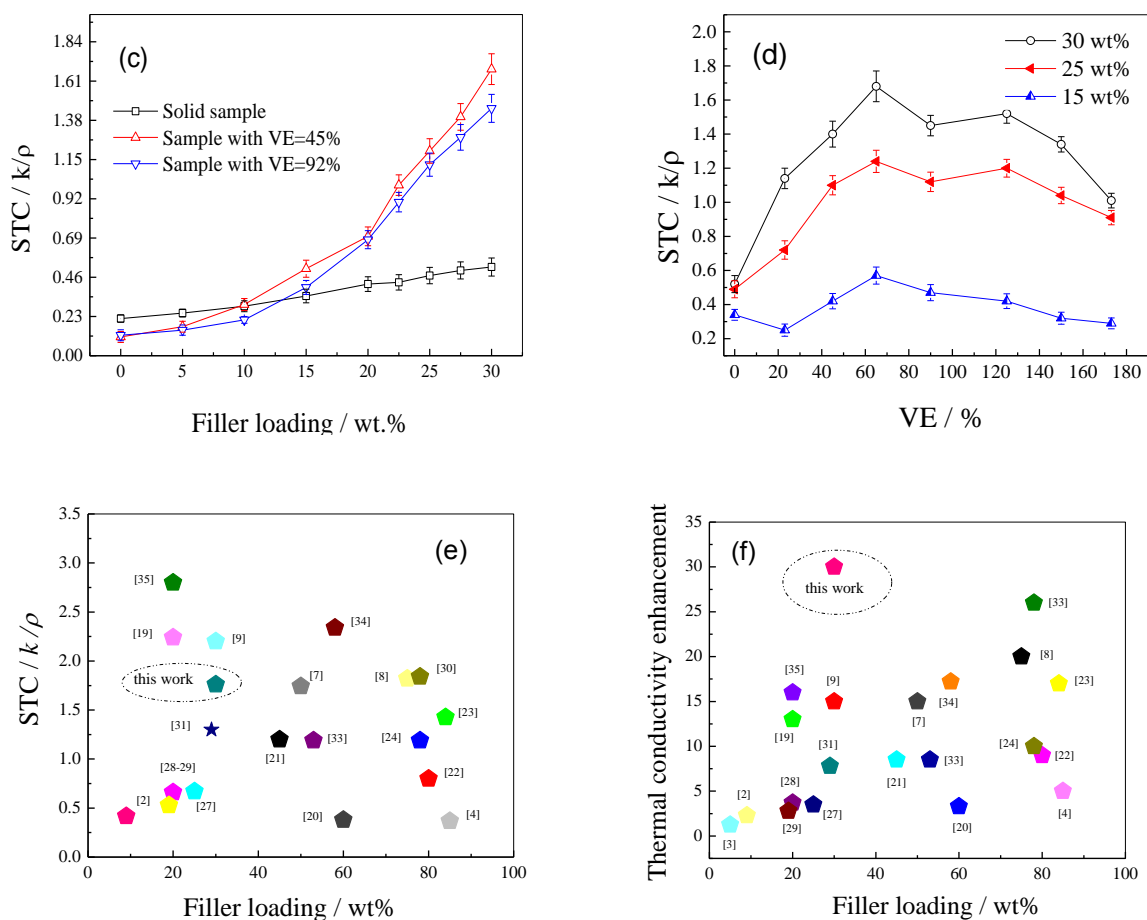


**Figure 4.** (a) Surface temperature variation with cooling time of pure c-PS (A), 30 wt% BNNSs/c-PS (VE=0) (B), 25 wt% BNNSs/c-PS (VE=45%) (C), and 30 wt% BNNSs/c-PS (VE=45%) (D); (b) Infrared thermal images of A, B, C and D at different times.

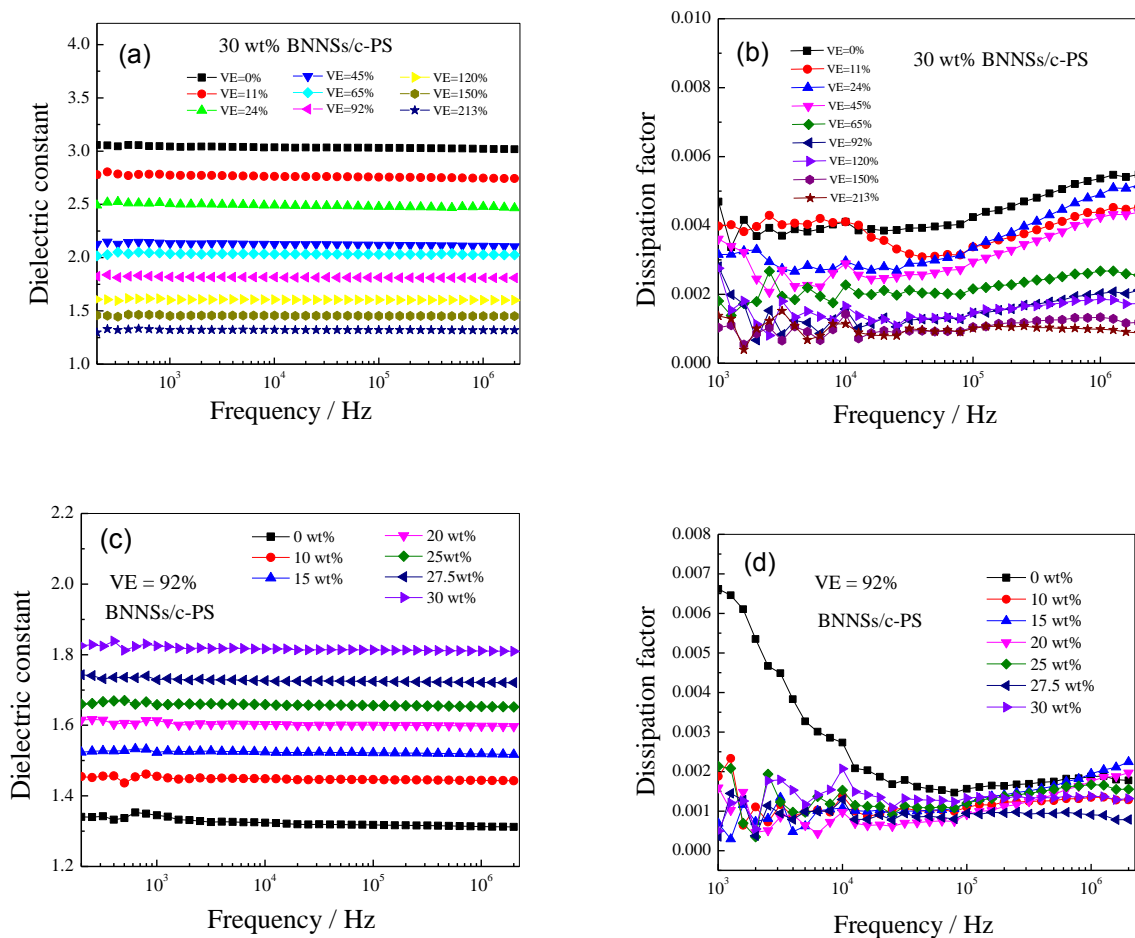


**Figure 5.** Micrographs of 30 wt% BNNs/c-PS (VE=0%) (a), 25 wt% BNNs/c-PS (VE=45%) (b-c), and 30 wt% BNNs/c-PS (VE=45%) (d), the inset is the dispersion of BNNs in c-PS matrix inside the cell wall.



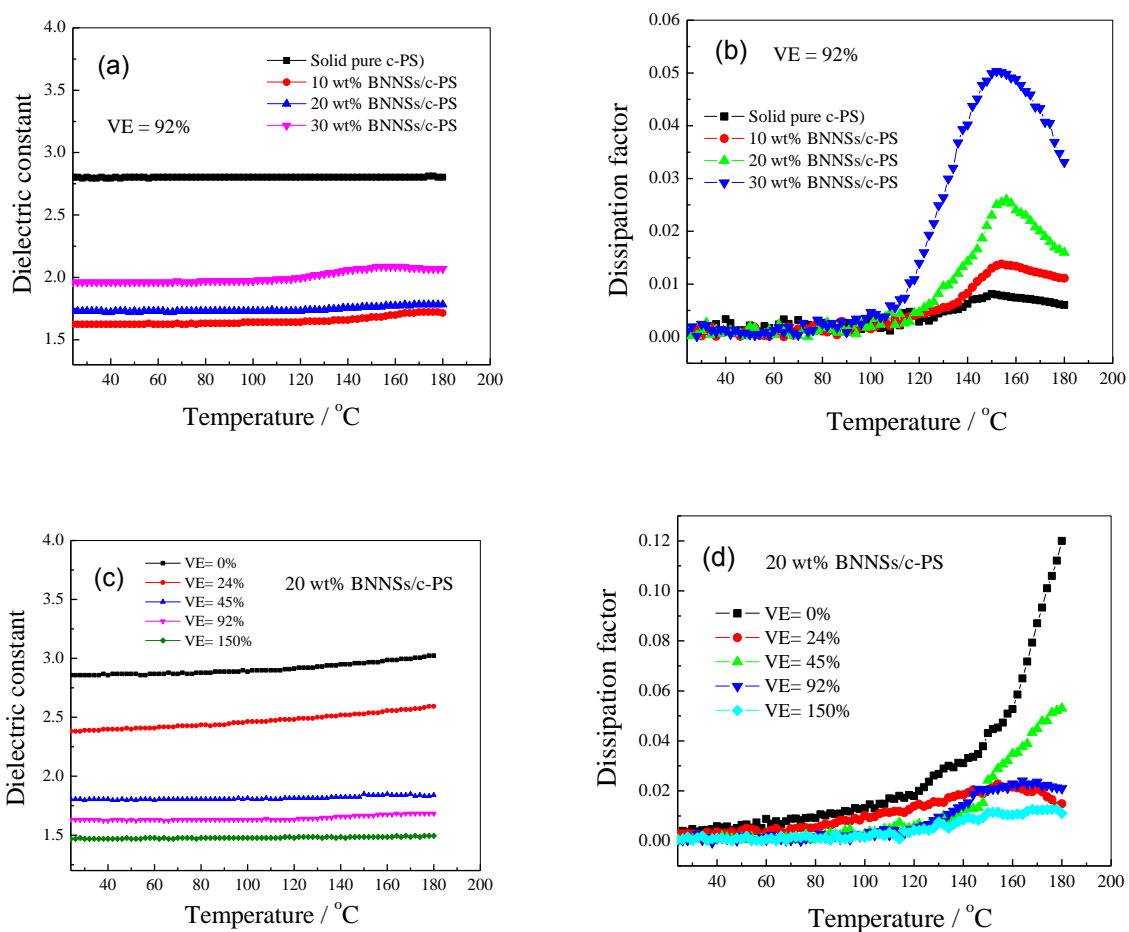


**Figure 6.** (a) Dependence of density of composite foams on VE, the inset photograph is a BNNSs/c-PS foam standing on a piece of tender branch, (b) Effects of VE on TC and density of composite foams with 25 wt% BNNSs, (c-d) STC as a function as filler loading and VE, (e) STC and (f) TC enhancement of composite foams and other composites in previous works.



**Figure 7.** (a) dielectric permittivity and (b) dissipation factor of 30 wt% BNNSSs/c-PS composites as a function of VE, (c) dielectric permittivity and (d) dissipation factor of BNNSSs/c-PS composites with a VE of 92% as a function of filler loading.





**Figure 8.** (a) dielectric permittivity and (b) dissipation factor (1000Hz) of c-PS composites with various loading of BNNSs at a VE of 92% as a function of filler loading, (c) dielectric permittivity and (d) dissipation factor (1000Hz) of 20 wt% BNNSs/c-PS composites as a function of VE.

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