Enhanced Thermoelectric Properties of Bi$_2$Te$_3$-Based Micro–Nano Fibers via Thermal Drawing and Interfacial Engineering

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High-performance thermoelectric (TE) materials with great flexibility and stability are urgently needed to efficiently convert heat energy into electrical power. Recently, intrinsically crystalline, mechanically stable, and flexible inorganic TE fibers that show TE properties comparable to their bulk counterparts have been of interest to researchers. Despite remarkable progress in moving TE fibers toward room-temperature TE conversion, the figure-of-merit value (ZT) and bending stability still need enhancement. Herein, interfacial-engineering-enhanced TE properties of micro–nano polycrystalline TE fibers fabricated by thermally drawing Bi$_2$Te$_3$-basedbulks in a glass-fiber template are reported. The interfacial engineering effect comes from generating stress-induced oriented nanocrystals to increase electrical conductivity and producing strain-distorted interfaces to decrease thermal conductivity. The 4 µm-diameter fibers achieve a 40% higher ZT (≈1.4 at 300 K) than their bulk counterparts and show a reversible bending radius of 50 µm, approaching the theoretical elastic limit. This fabrication strategy works for a wide range of inorganic TE materials and benefits the development of fiber-based micro-TE devices.

1. Introduction

As achieving a carbon dioxide emission peak and carbon neutrality are significant goals for alleviating global warming and energy problems, an effective solution is to recycle the distributed waste heat energy at medium-or-low temperatures (300–600 K). The capacities of direct conversion between heat energy and electricity have enabled thermoelectric (TE) devices to achieve power generation and solid-state cooling, in a zero-carbon and sustainable way.[1–3] The efficiency of TE materials is a key factor for the waste heat energy conversion, which is gauged by its dimensionless figure-of-merit value, $ZT = S^2\sigma T/\kappa$, where $S$ is the Seebeck coefficient, $\sigma$ is the electrical conductivity, $T$ is the absolute temperature, and $\kappa$ is the
thermal conductivity. However, the ZT value of commercial Bi$_2$Te$_3$-based materials is restricted to ~1 at room temperature owing to the adverse interdependence of the Seebeck coefficient, electrical conductivity, and thermal conductivity.[5,6] Since Bi$_2$Te$_3$-based bulk nanostructured materials have been demonstrated to be one of the best room-temperature TE materials with the highest ZT ≈ 1.4 at 300 K,[6–8] they possess large-scale application prospects for TE generators and refrigerators. For the Bi$_2$Te$_3$-based bulk materials, weak Van der Waals bonds between the layers reduce their mechanical flexibility. The elastic strain maxima of mechanically cut Bi$_2$Te$_3$-based pellets used in commercial TE devices were found to be 50% lower than its theoretical limit (5.5%).[9] Thus, the conventional TE devices can only offer limited flexibility, which makes them difficult to be used on small heat sources with curved shapes.[10–12]

Current solutions mainly focus on embedding/coating the Bi$_2$Te$_3$-based materials into/on flexible substrates such as yarns, textiles, films, paintings, and papers.[13–17] However, low TE performance, weak mechanical stability, and complex fabrication process still impede their applications. Hence, simultaneously improving ZT and flexibility of Bi$_2$Te$_3$-based TE materials is a great challenge, thus it is necessary to explore new strategies to meet the demand.

Over the past 50 years, optical fibers have been an extremely successful platform for optical materials and devices research. The internet and most telecommunication are delivered across the globe today through billions of kilometers of silica fibers, which are typically produced by thermally drawing a preform (Figure S2, Supporting Information). It shows that the average nanosheet thickness in the fiber core is ~20 nm. Furthermore, the fibers undergo rapid cooling with a rate of ~150 °C s$^{-1}$ right after thermal drawing, which should cause an interfacial defect as our reported Bi$_2$Te$_3$ fibers after preheating and fast drawing.[23] The defect illustration of the fiber cores is exhibited in Figure 1b, which shows that the carrier charges can quickly transport in the oriented nanosheets and multiscale defects can hinder phonon transport.

The TE properties parallel to the fiber-axis direction of the fibers were studied. Figure 1c,d shows the thermal conductivities (κ) and ZT values of the Bi$_2$Te$_3$ fibers (BTF), BST fibers (BSTF), and their corresponding bulks. The κ values of the BTF and BSTF are 50% lower than that of the corresponding bulks. Importantly, the BSTF with a diameter of 4 µm shows an ultralow κ of 0.58 W m$^{-1}$K$^{-1}$ and a high ZT of ≈1.4 at 300 K. The micro–nano fiber without glass cladding should possess a higher output power density than the reported mW cm$^{-2}$ level of TE fibers with glass cladding.[22] To highlight the advantages of the micro–nano TE fibers, the thermal conductivities and the ZT values of other state-of-the-art Bi$_2$Te$_3$-based fibers are also compared in Figure 1c,d.[22,23,29] The enhanced TE properties and the corresponding mechanism of the interfacial engineering effect will be discussed as follows.

2. Results and Discussion

2.1. Micro–Nano Thermoelectric Fibers Fabricated via Thermal Drawing

Figure 1a shows a schematic of thermal drawing Bi$_2$Te$_3$-based bulks into fibers. Micro–nano TE fibers with Bi$_2$Te$_3$ or BST core and borosilicate cladding were fabricated by thermal drawing preforms. Especially, the preforms have been endured 700 °C for a 2 h preheating time before drawing, to form a wetting interface between the core and the cladding by elemental diffusion (see Experimental Section; Figure S1, Supporting Information). The SEM and energy disperse spectroscope were carried out on the micro–nano fiber cross section with cladding and the hydrogen fluoride (HF) etched fiber core without cladding (Figure S2, Supporting Information). It shows that the average nanosheet thickness in the fiber core is ~20 nm. Furthermore, the fibers undergo rapid cooling with a rate of ~150 °C s$^{-1}$ right after thermal drawing, which should cause an interfacial defect as our reported Bi$_2$Te$_3$ fibers after preheating and fast drawing.[23] The defect illustration of the fiber cores is exhibited in Figure 1b, which shows that the carrier charges can quickly transport in the oriented nanosheets and multiscale defects can hinder phonon transport.

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2.2. Interfacial Microstructure

To understand the mechanism of the ultralow thermal conductivities and high electrical conductivities in the micro–nano fibers, further analysis was performed on the microfiber by EPMA, AFM, and TEM. The detailed information can be
found in Experimental Section. **Figure 2a,b** exhibits the linear elemental (Bi, Sb, Te, Si, O, and Na) profiles across the polished cross section of the BTF and BSTF by EPMA measurement. There exists little diffusion of Bi, Sb, and Te into the cladding, while there is some diffusion of Si, O, and Na into the core region. The diffused Na would enhance the carrier concentration and the electrical conductivity but reduce the Seebeck coefficient.\(^{[30]}\) The diffused O should form some nanoscale oxides in the core, such as SiO\(_2\) and TeO\(_2\), which will have strong phonons scattering, resulting in lower thermal conductivity as in our previous studies.\(^{[23]}\) The elemental diffusion introduces an oxidic surface of the fiber core, which could protect the fiber core from atmospheric oxidation.

**Figure 2c–f** shows the surface roughness of the BSTF cores with diameters of 100, 20, 5, and 1 \(\mu\)m, respectively. When the diameter is 100 \(\mu\)m, the surface of the as-prepared fiber core is relatively smooth, however, it becomes rougher as the core diameter decreases. When the diameter is 1 \(\mu\)m, the height of the humps dramatically increases. The surface roughness was measured to be 9, 33, 64, and 78 nm, corresponding to the fiber diameters of 100, 20, 5, and 1 \(\mu\)m, respectively. The enhancement in the surface roughness with the decreasing core diameter could be attributed to the increased depth of elemental diffusion as mentioned above. Notably, besides the well-recognized classical size effect due to the phonon boundary scattering, the surface roughness can also significantly alter the thermal conductivity.\(^{[31]}\)

In the fiber-confined field, we try to reveal the growth process of the oriented nanosheet crystal, whose crystal morphology and element distribution are shown in Figures S2–S4, Supporting Information. The morphology evolution during the fiber-drawing process is schematically shown in **Figure 3a**. During the cooling process of the first thermal drawing (part i), the Bi\(_2\)Te\(_3\) core melt gradually nucleates and grows into hexagonal nanosheets around the interface to minimize potential energy. The nanosheet crystals will continuously stack around the interface with stress and then the crystals grow along the (0 0 \(l\)) plane and compete in several directions into a flower shape (Figure S2b, Supporting Information). More interestingly, during the second drawing process (part ii), the nanosheet crystals stack around the interface at the beginning. Accompanied by smaller fiber-core space and larger interface stress than the first drawing process, then the crystals preferentially grow along the (0 0 \(l\)) plane in one direction (Figure S2c,
Supporting Information). It can be interpreted as a similar stress-induced structural orientation during a directional laser melting process or a hot pressing process. Meanwhile, elemental diffusion takes place between the cladding and the core, which causes structural defects and an increase in the surface roughness of the fiber core as found above.

Figure 3b presents a HAADF image of a BSTF sample processed by using a focused ion beam (FIB) method, including the upside core and the underside cladding of the fiber. Figure 3c–e shows the enlarged views of the circled region in Figure 3b, from the interface to the core of 0, 0.5, and 1 µm depth, respectively. The insects in Figure 3c–e present the corresponding images of selected-area electron diffraction (SAED). These measurements confirm the crystalline state of BST in the fiber core. Figure 3c shows that the core at the interface is in a nanoceramic state, which has a glass phase and a nanoscale polycrystalline phase, as reported in our previous study. The lattice-plane spacing of the polycrystal is 0.317 nm, corresponding to the (0 1 5) plane. Many polycrystals deformed in the white circled region, and their enlarged SAED patterns exhibit different dislocations and strain distortions (Figure S5, Supporting Information), which is similar to the reported distorted mosaic crystal with ultralow thermal conductivity. It is noted that the core at 0.5 µm depth away from the interface is in an ordered polycrystalline state, as shown in Figure 3d. And the insert of Figure 3d marks the corresponding lattice planes, including strong (0 0 15) and weak (3 0 0). Moreover, Figure 3e shows a single-crystal state with clear (3 0 0) planes, suggesting the direction of the BST (3 0 0) planes are parallel to the fiber axis. These results verify the effect of interfacial engineering on the ordered and defective microstructure of the fiber core (Figure 1b), which will hinder thermal transport more effectively than electrical transport.

2.3. Measured Room-Temperature Thermoelectric Properties

The electrical conductivity (σ), Seebeck coefficient (S), and thermal conductivity (κ) of the fiber were measured by using a pre-patterned silicon (Si) chip. Therefore the TE properties can be in situ measured from one fiber. As shown in Figure 4a, a Bi₂Te₃ fiber is placed over four golden (Au) electrodes for the thermal conductivity and electrical conductivity measurements on the test chip. Two outer electrodes (I⁺ and I⁻) and two
inner electrodes \((V_+\text{ and } V_-)\) are used for sourcing current and extracting voltage, respectively. One end of the fiber is located close to an Au heater which is used to create a temperature gradient for the \(S\) measurement. Once the heater was turned on, the two inner electrodes were also used as thermometers to probe the temperature difference across the central portion of the fiber. The bonding between the fiber and the electrodes was created by depositing platinum (Pt) by applying FIB. The chip was subsequently annealed at 40 \(^\circ\)C for 10 min to cure the welding spot of Pt. The axial thermal conductivity \(\kappa||\) of the suspended fiber was measured by the self-heating 3\(\omega\) method.\([28]\) In this method, the fiber is periodically heated by an alternating current (AC) at an angular frequency \(\omega\) (Figure S6, Supporting Information). The magnitude of the average temperature oscillation of the central portion of the fiber is proportional to a 3\(\omega\) voltage:

\[
|V_{3\omega}|_{\text{rms}} = \frac{4I_{\text{rms}}L^2R_s\beta_s}{\pi^4\kappa_A A_S(2\pi)^2} 
\]

where \(|V_{3\omega}|_{\text{rms}}\) and \(I_{\text{rms}}\) are the root-mean-square amplitude of the 3\(\omega\) voltage and the exciting current, respectively. \(L\) is the length of the central portion of the fiber. \(R_s\), \(\beta_s\), and \(A_S\) are the resistance, TCR, and the cross-sectional area, respectively. \(\gamma\) represents a characteristic thermal time constant associated with the heat transfer along the fiber axial direction. The measured \(|V_{3\omega}|_{\text{rms}}\) data can be used to fit Equation (1) to extract the axial thermal conductivity \(\kappa||\). Notice that a 30 \(\mu\)m-wide trench is created between the two inner electrodes \((V_+\text{ and } V_-)\) on the silicon substrate to make the central portion of the fiber (sensing region) suspended. This will eliminate the heat flow between the sensing region and the substrate, which is necessary for properly implementing the self-heating 3\(\omega\) method. Figure 4c shows the experimentally measured \(|V_{3\omega}|_{\text{rms}}\) values of the two fibers as a function of exciting current frequency \(f\) (\(f = \omega/2\pi\)). The obtained \(\kappa||\) of 3.5 \(\mu\)m-diameter BTF and 4 \(\mu\)m-diameter BSTF are 0.69 \(\pm\) 0.06 and 0.58 \(\pm\) 0.05 W m\(^{-1}\) K\(^{-1}\), respectively. Moreover, the glass cladding should affect the TE performance of the fiber core. In order to elucidate the cladding’s impact on the thermal conductivity, we carry out an SThM test\([35,36]\] on the cross section of BSTF as shown in Figure S7, Supporting Information. It shows that the average thermal conductivity of the glass cladding is 1.1 \(\pm\) 0.1 W m\(^{-1}\) K\(^{-1}\), while the average thermal conductivity of the fiber core is 1.08 \(\pm\) 0.1 W m\(^{-1}\) K\(^{-1}\), which is higher than that of the freestanding BSFT fiber core (0.58 W m\(^{-1}\) K\(^{-1}\)) determined by the 3\(\omega\) method. It can be attributed to the confined BSTF core in the cladding being under GPa-scale tensile stress,\([27]\) which could increase the electrical and thermal conductivity of the BSTF core by \(>20\%\). And the defects of oxides and rough surfaces on the freestanding fiber cores etched by HF should decrease their thermal conductivity either.

**Figure 3.** a) Oriented crystal growth model of the Bi\(_2\)Te\(_3\)-based fiber core during thermal drawing and cooling. b) HAADF image of a BSTF chip cut by Ga ion beam and c–e) enlarged view of the circled regions in (b). The insets of (c–e) are the corresponding SAED patterns.
The reliability of the current experimental configuration for measuring $\kappa_\parallel$ could be affected by the surface radiation of the fiber and the thermal contact of Pt bonding. To address these issues, we have applied the COMSOL software to perform a finite element modeling (FEM) calculation on a 4 $\mu$m-diameter fiber with a pre-set $\kappa_\parallel$ value of 0.5 W m$^{-1}$ K$^{-1}$ (Figure 4b). In the simulation, we assigned on purpose a unity emissivity to the surfaces of the fiber and Pt bonding. The thermophysical properties and experimental dimension of the Pt bonding were also carefully considered. Figure 4d shows the theoretically predicted peak amplitude of temperature oscillation for the 4 $\mu$m-diameter fiber. By performing a similar nonlinear fitting on the data shown in Figure 4d (see Experimental Section), the $\kappa_\parallel$ derived from the FEM simulation is found to be 0.455 W m$^{-1}$ K$^{-1}$, which is about 9% smaller than its pre-set value. Thus, we believe that the $3\omega$ measurements based on the current configuration are reliable. The temperature-dependent electrical resistances of two fibers were measured by the four-probe method, from which the TCRs in Equation (1) and the $\sigma$ can be calculated (Figure 4e). The $S$ values of the two fibers were measured by using a steady-state method.$^{[38]}$ In this method, the heater was powered by a DC current source, which can create temperature gradients of different magnitude across the fiber, and the corresponding TE voltages ($dV$) were measured by a Keithley 2400 sourcemeter (Figure 4f). The $S$ can be calculated from the slope of the $dV$–$dT$ linear fitting. Detailed descriptions of the experimental implementation and simulation of the self-heating $3\omega$ method can be found in Experimental Section.

The obtained TE properties of the fiber samples are listed in Table 1 with measuring uncertainties (Note S1, Supporting
Information) in comparison with those reported in the literature. Except for $\Sigma$, the values of $\sigma$, $\kappa|_l$, and $ZT$ of the prepared fibers are all improved with respect to those of bulk materials at different levels. Compared with the other reported TE fibers, the fiber core exhibits moderate $S$, $\sigma$, and low $\kappa|_l$ values, which are mainly caused by the interfacial defect structure derived from elemental diffusion during the twice thermal drawing processes. The nanosheet thickness of $\approx$20 nm is close to the wavelength of phonons and larger than that of majority carriers, so the nanosheet boundaries could selectively scatter phonons and make carriers pass through.[39] The phonon modulation is then investigated by Vienna Ab initio Simulation Package (VASP) software based on the density functional theory. The calculation details can be found in Note S2, Supporting Information. The study confirms the effect of different mean-free-path phonons on the lattice thermal conductivity of Bi$_2$Te$_3$ at 300 K (Figure S8, Supporting Information). What’s more, the nanostructure of Bi$_2$Te$_3$ under 20 nm could efficiently scatter phonons and depress the lattice thermal conductivity at 300 K. The thermal conductivity along the whole fiber should be estimated as $\kappa|_l = \kappa|_l + \kappa|_b + \kappa|_s$, where $\kappa|_l$ is the lattice thermal conductivity, $\kappa|_e$ is the electronic thermal conductivity, and $\kappa|_s$ is the bipolar thermal conductivity. When the $\kappa|_b$ might be negligible at 300 K,[40] the electronic thermal conductivity then can be estimated and found in Note S3, Supporting Information.

Beyond these, the bipolar transport is important for Bi$_2$Te$_3$-based TE materials at a temperature higher than 300 K. The thermal conductivity along the whole fiber should be estimated and found in Note S3, Supporting Information. The maximum output power of the TE fiber generator under 2.6 K are summarized in Figure S11, Supporting Information. The maximum output power of the TE fiber generator by varying the load resistance. The fiber system investigated in this study may provide another opportunity to address this issue in the low-dimensional system in the future.

As shown in Table 1, the electrical conductivities of the fibers are $\approx$20% higher than those of the bulks, which can be attributed to the enhanced carrier mobility of the stress-induced oriented nanocrystals as found in our previous studies.[21,27] The fiber core without cladding can keep 90% of its original electrical conductivity value after being preserved in the atmosphere for more than 1 year. (Figure S9, Supporting Information). And these fibers do not require using ion milling/etching as reported for nanowires to cleave surface oxides and measure TE properties.[29,41] Beyond these, we carry out a bending procedure and consecutively measure I–V characteristics of the representative BTF and BSTF (fiber diameter $D = 4$ $\mu$m) by applying two gold-plated tungsten probes and silver paste (Figure 4g; Figure S10, Supporting Information). Using the original resistance $R_0$ as a reference, the bending-state resistance $R$ is normalized to exhibit a change of the electrical resistance as a function of 0.05–5 mm bending radius $r$ (Figure 4h). Within $\approx$5% measuring uncertainty, the resistance of both samples increases with decreasing bending radius. The variation of the resistance is reversible, which should be caused by the bending flexibility of the (0 0 1)-orientation layer structure.[13]

After exerting 100-time bending cycles at $r = 50$ $\mu$m, the resistance increases by less than 10%, indicating its tremendous mechanical flexibility (Figure S10, Movie S1, Supporting Information). The bending radius minima ($r_{\text{min}}$) of all referred samples are also listed in Table 1. Their flexibility can also be estimated by the maximum bending strain ($\varepsilon = D/2r$).[42] It exhibits that these microfibers possess $r_{\text{min}} = 50$ $\mu$m as an excellent $\varepsilon$ of 4%, approaching that of the reported nanowires and the $\approx$5.5% theoretical limit of Bi$_2$Te$_3$ elastic strain.[9,43] These results demonstrate that the thermally drawn micro–nano Bi$_2$Te$_3$ fibers with interfacial engineering effects possess both high TE performance and great bending stability.

To demonstrate the TE properties, our microdevice can be viewed as a prototype of a single-legged TE generator (with 4 $\mu$m-diameter BTSF and 30 $\mu$m-length suspended span) to measure the output performance. Under a stable temperature difference of $\approx$2.6 K across the TE fiber, the device exhibited an open-circuit voltage of 377 $\mu$V and a short circuit current of 4.24 $\mu$A. We then measured the output current and output voltage of the TE fiber generator by varying the load resistance. The output current, output voltages, and output power of the TE fiber generator under 2.6 K are summarized in Figure S11, Supporting Information. The maximum output power of $\approx$0.4 nW was obtained for this device. Thus the single fiber-based generator achieves a maximum output power density (output power per unit area) up to 3.18 mW cm$^{-2}$ under a temperature difference of 2.6 K, which is 36% higher than that of the reported 7-pairs of p–n fibers in the TE device under a temperature difference of 60 K.[25]

### Table 1. Thermoelectric and bending properties of Bi$_2$Te$_3$ bulk and fiber samples.

| Samples at 300 K$^s$ | $S$ [µV K$^{-1}$] | $\Sigma$ [$\times 10^5$ S m$^{-1}$] | $\kappa$ [W m$^{-1}$K$^{-1}$] | ZT | $r_{\text{min}}$ |
|----------------------|-----------------|---------------------------------|-----------------|----|----------------|
| Bulk Bi$_2$Te$_3$    | 114             | 5.0                             | 1.50            | 0.21 | /              |
| 3.5 $\mu$m BTF      | 107 ± 12        | 5.9 ± 0.3                       | 0.69 ± 0.06     | 0.3 ± 0.07 | 45 ± 5 $\mu$m |
| Bulk BST            | 212             | 10.5                            | 1.4             | 1.01 | /              |
| 4 $\mu$m BSTF       | 145 ± 16        | 12.5 ± 0.7                      | 0.58 ± 0.05     | 1.38 ± 0.34 | 50 ± 5 $\mu$m |
| 30 $\mu$m BSTF      | 150             | 15.6                            | 0.84            | 1.25 | 1 cm           |
| 50 $\mu$m Bi$_2$Te$_3$ | 131          | 7.4                             | 0.52            | 0.73 | 5 cm           |
| 65 nm BSTF$^f$      | 125             | 4.5                             | 0.3             | 0.7  | 500 nm         |

$^s$BTF: Bi$_2$Te$_3$ fibers, BST: Bi$_{0.5}$Sb$_{1.5}$Te$_3$, BSTF: Bi$_{0.5}$Sb$_{1.5}$Te$_3$ fibers.

### 3. Conclusions

We have demonstrated the high-performance micro–nano polycrystalline Bi$_2$Te$_3$-based TE fibers with great flexibility and stability through thermal drawing Bi$_2$Te$_3$-based bulks in a
glass-fiber template. A special interface between the glass cladding and the Bi₂Te₃-based core was observed to induce oriented nanosheet crystals in the fiber cores, enabling the fibers to display ultralow thermal conductivity and high electrical conductivity. Furthermore, this interfacial engineering efficiently regulates the electrical and thermal transport, while keeping the electrical properties stable after 100-times 50 µm-diameter bending and long-time preservation. This strategy could be universal to other inorganic TE materials, particularly those made from nontoxic and abundant elements. The high TE performance and stability in the micro–nano TE fibers will stimulate further research on the TE fiber materials (e.g., fibrous crystal anisotropy, quantum size effect, and bipolar transport in the 1D system) and the development of integrated TE fiber devices (e.g., self-powered wearable electronics and micro-area temperature management).

4. Experimental Section

Fabrication of Micro–Nano TE Fibers: The glass-fiber-template method and two-step thermal drawing were applied for micro–nano Bi₂Te₃ fibers fabrication. First, a 10 mm-diameter Bi₂Te₃ or Bi₀.5Sb₁.5Te₃ (BST) rod (99.99% purity, Santech Materials, Changsha) was sealed in a Pyrex 3.3-

Composition and Microstructure Characterizations: Utilizing an argon ion beam cross section polisher (CP, JEOL IB-09020CP, Japan), the as-drawn fibers were polished to obtain a cross-sectional sample and a longitudinal section sample. And the two fiber samples were observed under a field-emission scanning electron microscope (ZEISS Merlin, Germany) operated at 20 kV with a working distance of 8.4 mm. The Bi₂Te₃ core was controlled to be within ±1% when the fiber diameter ranged from 9.5 to 475 µm. As the ratio of core–cladding diameter was about 1.95 when the fiber core diameter ranged from 0.1 to 5 µm, theoretically the diameter fluctuation of the fiber core could be controlled to be within ±1%.

Finite Element Simulation of the 3ω Measurement: A 3D FEM simulation was carried out to evaluate the reliability of measuring the thermal conductivity of the fiber on the test chip with the geometrical configuration shown in Figure 4a. A simulation model was constructed to imitate the chip setup, in which the diameter and length of the suspended portion of the fiber were 4 and 30 µm, respectively. (Figure 4b) The fiber was assigned a thermal conductivity of 0.5 W m⁻¹ K⁻¹ and a volumetric heat capacity of 1.22 × 10⁶ J m⁻³ K⁻¹. Lee and co-workers found that the platinum film created by the FIB deposition mainly consisted of carbon (45–55%) and platinum (40–50%).[47] The thermal conductivity of the Pt/C film was measured to be 40.97 W m⁻¹ K⁻¹. In the simulation, the thermal conductivity of the Pt bonding was set to be 40 W m⁻¹ K⁻¹. It was further assumed that the Pt/C bonding was composed of 50% amorphous carbon and 50% platinum in volume.[47] Therefore the volumetric heat capacity of the Pt/C bonding was 98 × 10⁶ J m⁻³ K⁻¹. From Lee’s work,[48] it was also estimated that the electrical conductivity of the Pt/C material was about 72.04 S m⁻¹. To initiate the simulation, an AC electrical current with a peak amplitude iₒ of 0.4 mA and an angular frequency of ω was injected into the fiber from two outer electrodes to create periodical heating at frequency 2ω. Radiation loss was modeled by employing the Stefan–Boltzmann law.

Table 2. Materials properties used in the FEM simulation.

|                  | Bi₂Te₃[44] | Pt/C[44] | Au[45] | Si[45] |
|------------------|-----------|----------|--------|--------|
| k × 10⁻³ W m⁻¹ K⁻¹ | 0.5       | 40       | 240    | 142.2  |
| Cᵥ J m⁻³ K⁻¹     | 1.22 × 10⁶ | 2.98 × 10⁶ | 1.60 × 10⁶ | 1.63 × 10⁶ |
| γ                 | 1         | 1        | 0.03   | 0.5    |
| σ S m⁻¹           | 1.25 × 10⁶ | 72.4     | 5.23 × 10⁶ | /      |
\[ f = 15 \text{ Hz} \ (f = \omega / 2\pi). \] The peak amplitude \(|\Delta T_{2\omega}|_p\) of the temperature oscillation could be displayed as a function of exciting current frequency, as shown in Figure 4d. Notice that the real experiment \(|\Delta T_{2\omega}|_p\) could be derived from the root-mean-square amplitude of the 3ω voltage \(|V_{3\omega}|_{rms}\) by:

\[ |V_{3\omega}|_{rms} = \frac{l_{rms} R_s \beta_s |\Delta T_{2\omega}|_p}{2} \tag{2} \]

where \(l_{rms}\) is the root-mean-square amplitude of the exciting current. \(R_s\) is the resistance of the suspended portion of the fiber. \(\beta_s\) is the TCR of the fiber. Substituting Equation (2) into Equation (1) yields temperature oscillation:

\[ |\Delta T_{2\omega}|_p = \frac{2l_{rms} R_s L}{\pi \kappa \omega^2} (1 + (2\omega l)^2) \tag{3} \]

where \(A_s\) is the cross-sectional area of the fiber. By fitting the \(|\Delta T_{2\omega}|_p\) versus \(f\) curve in Figure 4d to Equation (3), the thermal conductivity was found to be 0.455 W m\(^{-1}\) K\(^{-1}\).

**Statistical Analysis:** The experimental characterizations such as XRD, SEM, and electroprobe microanalyzer (EPMA) were carried out at least three samples. The TEM, atomic force microscopy (AFM), and SThM measurements were performed on the same sample more than three times. All results showed good consistency. The TE transport properties and bending resistances were measured on the same sample three times to yield averaged results. More detailed description on error analysis on TE properties measurements can be found in Supporting Information.

**Supporting Information**

Supporting Information is available from the Wiley Online Library or from the author.

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**Conflict of Interest**

The authors declare no conflict of interest.

**Author Contributions**

M.S., G.W.T., and H.F.W. contributed equally to this work. M.S., G.W.T., P.Y.Z., and B.H. prepared the micro–nano TE fibers and conducted SEM, EPMA, and TEM characterization. M.S., G.W.T., and H.F.W. performed TE properties measurements and wrote the original manuscript. M.Y., H.Z., and T.Z. performed theoretical simulation. M.S., Y.C.C., and J.C. conducted stability tests of electrical performance. Q.F.Z. and J.Y.L. conducted SThM of thermal conductivity distribution. M.S., G.W.T., H.F.W., D.C.C., and J.L.G. analyzed the results. Q.Q. and Z.M.Y. directed this project.

**Data Availability Statement**

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

**Keywords**

Bi\(_2\)Te\(_3\), interfacial engineering, thermal drawing, thermoelectric fibers

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