Room temperature magnetization switching in topological insulator-ferromagnet heterostructures by spin-orbit torques

Yi Wang¹, Dapeng Zhu¹, Yang Wu¹, Yumeng Yang¹, Jiawei Yu¹, Rajagopalan Ramaswamy¹, Rahul Mishra¹, Shuyuan Shi¹, Mehrdad Elyasi¹, Kie-Leong Teo¹, Yihong Wu¹ & Hyunsoo Yang¹,²

Topological insulators with spin-momentum-locked topological surface states are expected to exhibit a giant spin-orbit torque in the topological insulator/ferromagnet systems. To date, the topological insulator spin-orbit torque-driven magnetization switching is solely reported in a Cr-doped topological insulator at 1.9 K. Here we directly show giant spin-orbit torque-driven magnetization switching in a Bi₂Se₃/NiFe heterostructure at room temperature captured using a magneto-optic Kerr effect microscope. We identify a large charge-to-spin conversion efficiency of -1-1.75 in the thin Bi₂Se₃ films, where the topological surface states are dominant. In addition, we find the current density required for the magnetization switching is extremely low, ~6 × 10⁵ A cm⁻², which is one to two orders of magnitude smaller than that with heavy metals. Our demonstration of room temperature magnetization switching of a conventional 3d ferromagnet using Bi₂Se₃ may lead to potential innovations in topological insulator-based spintronic applications.
The spin currents generated by charge currents via the spin Hall effect and/or Rashba-Edelstein effect can exert spin-orbit torques (SOTs) on the adjacent FM layer and result in the current-induced magnetization switching. A higher charge-to-spin conversion efficiency (referred as SOT efficiency) is crucial for the low-power dissipation SOT applications. Recently, the SOTs have been studied in topological insulators (TIs), which are an emerging state of quantum matter possessing spin-momentum-locked topological surface states (TSSs). This exotic property is supposed to exhibit a large SOT efficiency, which is explored recently by the spin transport methods such as spin-torque ferromagnetic resonance (ST-FMR), spin pumping, and spin tunneling spectroscopy. However, in TIs such as Bi$_2$Se$_3$, the bulk states (BS) and two-dimensional electron gas (2DEG), which are typically present due to defects in the bulk and band bending at the surface, respectively, can lead to an inevitable contamination in the SOT effects from TSS. This is indicated by a wide range of the SOT efficiencies of 0.01–3.5 reported in the Bi$_2$Se$_3$/ferromagnet (FM) systems and TSS on efficiencies have not yet been clearly understood in details, which is critical for highly efficient SOT-driven magnetization switching using TIs.

To date, the magnetization switching induced by TI SOT is solely reported in a Cr-doped TI at a very low temperature (1.9 K) with an external magnetic field, and the SOT-induced magnetization switching in a TI/3d FM heterostructure at room temperature is highly desired for applications. Here we obtain a TSS–dominated SOT effect in 5–8 quintuple layers (QL) of Bi$_2$Se$_3$ films, exhibiting a large SOT efficiency of ~1–1.75 at room temperature using ST-FMR measurements. By taking advantage of the high efficiency, we image the SOT-induced magnetization switching by a magneto-optic Kerr effect (MOKE) microscope in the Bi$_2$Se$_3$/NiFe (Py) heterostructures at room temperature after injecting a pulsed dc current. The required current density for SOT switching is extremely low and is one to two orders of magnitude smaller than that with heavy metals. Our results suggest that TI/FM heterostructure could be a potential candidate for room temperature spintronic devices with ultralow-power dissipation.

**Results**

**Bi$_2$Se$_3$ growth and film characterization.** High-quality Bi$_2$Se$_3$ films ranging from 5 to 20 QL (1 QL ≈ 1 nm) were grown on Al$_2$O$_3$ (0001) substrates using molecular beam epitaxy (MBE) technique (see Methods). Figure 1a shows the atomic-force microscopy (AFM) image of a representative 10-QL Bi$_2$Se$_3$ film, indicating a smooth surface and high film quality. From the four-probe and Hall measurements at room temperature, we find that the resistivity ($\rho_{\text{BS}}$) is ~1,000 μΩ cm at large thicknesses (15 and 20 QL), increases at 10 QL and becomes ~4,117 μΩ cm at 5 QL, as shown in Fig. 1b. Moreover, the sheet resistance shows a similar trend as $\rho_{\text{BS}}$ (Supplementary Note 2 and Supplementary Fig. 2). The sheet carrier concentration ($n_{2\text{D}}$) shows an opposite trend, decreasing from ~6 × 10$^{13}$ cm$^{-2}$ at 20 QL to ~3.8 × 10$^{13}$ cm$^{-2}$ below 8 QL. This behavior suggests a small contribution of BS and 2DEG to electrical transport properties in the thin Bi$_2$Se$_3$ cases, as we discuss later. We also characterize Bi$_2$Se$_3$ thickness ($t_{\text{BS}}$)-dependent $\rho_{\text{BS}}$ (Fig. 1c) and $n_{2\text{D}}$ (Fig. 1d) at different temperatures. Our Bi$_2$Se$_3$ films show a typical metallic behavior similar to previous reports.

**ST-FMR measurements.** Figure 2a shows the schematic diagram of the ST-FMR measurement (see Methods section), an effective technique to evaluate the SOT efficiency. The ST-FMR devices consist of Bi$_2$Se$_3$ ($t_{\text{BS}}$)/Co$_{40}$Fe$_{40}$B$_{20}$ (CFB, 7 nm) bilayers. Figure 2b illustrates the current-induced spin polarization and magnetization dynamics in Bi$_2$Se$_3$/CFB bilayers. As an in-plane rf current ($I_{\text{RF}}$) flows in the Bi$_2$Se$_3$ layer, non-equilibrium spins are...
generated at the Bi$_2$Se$_3$ surfaces denoted by the arrows with green and red balls. These spins from Bi$_2$Se$_3$ top surface diffuse into CFB and exert oscillating damping-like torque ($\tau_D$) and/or a field-like torque ($\tau_F$) on the magnetization. These torques together with rf current-induced Oersted field ($H_{RF}$) trigger the precession of CFB magnetization and an oscillation of the anisotropic magnetoresistance with the same frequency as $H_{RF}$. Consequently, a mixing dc voltage $V_{mix}$ (i.e., ST-FMR signal) is produced across the ST-FMR device.\(^7,8,27,28\)

Figure 2c shows typical ST-FMR signals $V_{mix}$ (open symbols), which are fitted by $V_{mix} = V_{S}F_{S} + V_{A}F_{A}$, where $F_{S}$ and $F_{A}$ are symmetric and antisymmetric Lorentzian functions, respectively. The amplitudes of symmetric ($V_{S}$) and antisymmetric component ($V_{A}$) are attributed to $\tau_{DL}$ and $\tau_{FL} + \tau_{BO}$, respectively.\(^22,27\) By adopting the established analysis method,\(^7,8\) the SOT efficiency ($\theta_{TL} = J_{S}/J_{C}$) can be evaluated from only $V_S$ (Supplementary Note 3), where $J_{S}$ is the spin current density at the Bi$_2$Se$_3$/CFB interface and $J_{C}$ (A cm$^{-2}$) is the uniform charge current density in the Bi$_2$Se$_3$ layer. Similar measurements and analyses are further performed on devices with various $t_{BiSe}$ spanning 5–20 QL. Figure 3 shows $\theta_{TL}$ vs. $t_{BiSe}$ at room temperature. Each data point is averaged from three devices, which show a similar behavior. Specifically, $\theta_{TL}$ shows a constant value of ~0.3 for 15 and 20 QL devices, and starts to increase below 10 QL, reaching a maximum of ~1.75 at 5 QL. $\theta_{TL}$ in thinner films increases ~5 times compared to that in thicker Bi$_2$Se$_3$ devices. From the line shape of the ST-FMR signals and the positive sign of $\theta_{TL}$ at different $t_{BiSe}$, we confirm that the direction of in-plane spin polarization ($S_{||}$) at the interface of Bi$_2$Se$_3$ and CFB is in line with TSS where $S_{||}$ is locked at right angles to the electron momentum.\(^7,8,29,30\)

### SOT efficiency vs. Bi$_2$Se$_3$ thickness and the role of TSS

Recent experimental and theoretical works\(^21,32\) indicate that the thickness of Bi$_2$Se$_3$ decreases to several QLs, BS shrink significantly and finally disappear. In addition, the surface 2DEG bands are gradually quantized into discrete subbands enclosed by the linear TSS bands due to quantum confinement effects (Supplementary Fig. 3). Since the thickness of a TSS ($t_{TSS}$) and 2DEG ($t_{2DEG}$) in Bi$_2$Se$_3$ are reported to be ~1 nm\(^21,26,32,33\) and ~4 nm\(^21,26,32\), respectively, negligible BS are expected when the

The amplitudes of symmetric ($V_{S}$) and antisymmetric Lorentzian ($V_{A}$) components, respectively.

The SOT efficiency ($\theta_{TL}$) is a function of Bi$_2$Se$_3$ thickness ($t_{BiSe}$) at room temperature. Each $\theta_{TL}$ represents the averaged value from three devices. The error bars are the standard deviation. Region I, II, and III denoted by different colors represent the charge-to-spin conversion dominated by different mechanisms. The inset shows the schematic of the band structure for each region. Bi$_2$Se$_3$ thickness is < 8 QL. Accordingly, we discuss the transports in three regions (I, II, and III, denoted by different colors) in Fig. 3. In region I ($t_{BiSe}$ > 10 QL), there are considerable BS and 2DEG contributions to the transport, which could dilute the TSS\(^9\) resulting in a small $\theta_{TL}$. In region II (~10 QL), BS start to shrink, leading to a slight increase of $\theta_{TL}$. In region III ($t_{BiSe}$ ≤ 8 QL), the BS disappear and the contribution from the 2DEG decreases as we discuss later. On the other hand, due to the lack of inversion symmetry in our devices, Rashba splitting states in 2DEG subbands can rise to $S_{||}$. However, the accumulated spins due to the Rashba states are expected to have an opposite helicity (i.e., negative $\theta_{TL}$) compared to the TSS\(^21,34-36\). Since $\theta_{TL}$ always shows positive values in all our devices, we conclude that the TSS dominated SOT is the main contribution to the large enhancement of $\theta_{TL}$ in region III.

To further confirm that the TSS dominate SOT in region III (5–8 QL), we establish a model to quantify the carrier concentration in TSS ($n_{TSS}$), 2DEG ($n_{2DEG}$) and BS ($n_{2D-Bulk}$), as well as the corresponding current shunting effect due to BS and 2DEG (Supplementary Note 4 and Supplementary Fig. 3). As shown in Fig. 4a, the $n_{2DEG}$ decreases significantly as $t_{BiSe}$ ≤ 8 QL, while $n_{TSS}$ shows a slight increase as $t_{BiSe}$ increases. This observation reproduces the inherent behaviors of TSS and 2DEG carriers measured in very thin Bi$_2$Se$_3$ films.\(^26\) Moreover, the larger value of $n_{TSS}$ compared to $n_{2DEG}$ for $t_{BiSe}$ ≤ 8 QL corroborates a TSS dominated transport in thin Bi$_2$Se$_3$ film region (Supplementary Note 4 and Supplementary Figs. 4–7). Figure 4b shows the location of the Fermi level ($E_{F}$) relative to the Dirac point ($E_{DP}$) and Fermi vector ($k_{F}$), we find that $E_{F} - E_{DP}$ ($k_{F}$) gradually increases from ~403 to 447 meV (from ~0.123 to 0.135 Å$^{-1}$) as $t_{BiSe}$ decreases, indicating that the DP slightly moves downwards to a larger binding energy which accounts for the weak increase of $n_{TSS}$. The value of $E_{F} - E_{DP}$ ($k_{F}$) and the DP movement are in line with previous ARPES measurements for Bi$_2$Se$_3$ films.\(^35,37-40\) Figure 4c shows that the charge currents in the TSS on the top surface over the total currents flowing in Bi$_2$Se$_3$ ($I_{TSS}/I_{total}$) increases from ~0.2 to 0.4 as $t_{BiSe}$ decreases from 20 to 5 QL (Supplementary Note 5), which again verifies that the TSS dominates the region III.

In addition, we estimate the “interface” SOT efficiency from TSS ($\lambda_{TSS}$) by using an interface charge current density $J_{C,TSS}$ (A cm$^{-2}$) in TSS (Supplementary Note 6 and Supplementary Figs. 8–11). As shown in Fig. 4d, we find $\lambda_{TSS}$ is in the range of ~0.38 to 0.82 nm$^{-1}$ when $t_{BiSe}$ ≤ 8 QL at room temperature, which is consistent with recently reported interface SOT efficiency.
values in \((\text{Bi}_{1-x}\text{Sb})_x\text{Te}_3\). In principle, \(\lambda_{\text{TSS}}\) is inversely proportional to the Fermi velocity \(V_F\) and remains almost constant at different \(t_{\text{BiSe}}\), however, a pronounced variation of \(\lambda_{\text{TSS}}\) is observed. This deviation unambiguously suggests that there is an opposite spin accumulation mechanism which cancels part of the spins generated by TSS in Bi2Se3. We attribute this to the Rashba states in 2DEG. From the change of \(\lambda_{\text{TSS}}\), we can extract the interface SOT efficiency from 2DEG (\(\lambda_{\text{2DEG}}\)) \(\approx 0.4\) nm\(^{-1}\) in the thin film regime (Supplementary Note 7). After excluding the 2DEG contribution, the amended interface SOT efficiency from TSS denoted by red circles in Fig. 4d, shows a constant value of \(-0.8\) nm\(^{-1}\) for 7, 8 and 10-QL Bi2Se3 devices, which is similar to the value of \(\lambda_{\text{TSS}}\) \(\approx 0.82\) nm\(^{-1}\) at \(t_{\text{BiSe}} = 5\) QL (Supplementary Note 7 and Supplementary Figs. 8–11).

**SOT-driven magnetization switching in Bi2Se3/Py.** The ST-FMR measurements and the above analysis reveal that the contribution of TSS is dominant in the thin Bi2Se3 films (5–8 QL), leading to a higher SOT efficiency at room temperature. Subsequently, we demonstrate the SOT-induced magnetization switching in Bi2Se3 (8 QL)/Py (6 nm) heterostructures (see Methods section) at room temperature by applying a pulsed current \(I\). The high-resolution scanning transmission electron microscope (STEM) image shows a clean and smooth interface between the Bi2Se3 and Py layer (see Supplementary Note 1 and Supplementary Fig. 1). In order to take advantage of the higher SOT efficiency and flow enough charge currents in the Bi2Se3 layer, an 8-QL Bi2Se3 is utilized. As depicted in Fig. 5a, the continuous Py layer is separated into five well-defined rectangles (yellow dashed boxes) and magnetically isolated by Cu bars. The magnetic easy axis of Py rectangles is along \(\pm y\) directions due to the shape anisotropy. The magnetization direction of Py is collinear with the incoming spin directions (Fig. 2b) and thus the spins can directly switch the magnetization direction of Py without any external assisted magnetic field, which are captured by MOKE imaging measurements (see Methods section).

Figure 5a–e (top panel) show the SOT-driven magnetization switching by applying a pulsed \(I\) along the \(+x\)-axis. At the beginning of this set of measurements, we first saturate the Py magnetization along the \(+y\)-axis by applying an in-plane external magnetic field \((H)\). Then, we remove \(H\) and apply \(-y\)-axis to the device. When the current density in Bi2Se3 \((J_C)\) is zero, we capture the MOKE image as shown in Fig. 5a. The dark contrast represents the magnetization along the \(+y\)-axis, indicated by the white arrow. We find that as \(J_C\) increases, the area of the switched magnetization with light contrast gradually expands (see Fig. 5b–d). Finally, the magnetization of all Py rectangles is switched to the \(-y\)-axis at \(J_C = 5.7 \times 10^5\) A cm\(^{-2}\), which is indicated by the white arrow in Fig. 5e. Similarly, for the other set of measurements in Fig. 5f–j, we first initialize the Py magnetization along the \(-y\)-axis. Then we remove \(H\) and apply \(I\) of opposite polarity, i.e., along the \(-x\)-axis. As \(J_C\) increases, the Py magnetization switches from the \(-y\) (Fig. 5f, light contrast) to \(+y\)-axis (Fig. 5j, dark contrast) at \(J_C = 6.2 \times 10^5\) A cm\(^{-2}\), exhibiting the opposite switching direction. The SOT-induced switching is reproducible in other devices (Supplementary Note 8 and Supplementary Fig. 12). We find that the current density required for the room temperature SOT-induced magnetization switching in Bi2Se3/Py is extremely low at \(-6 \times 10^5\) A cm\(^{-2}\), which is one to two orders of magnitude smaller than that with heavy metals\(^{23–25}\). Moreover, based on the conventional antidamping spin-torque-driven magnetization switching model\(^{24,41}\) with consideration of thermal fluctuation and reverse domain nucleation, we determine the SOT efficiency for Bi2Se3/Py to be \(-1.71\), which is in accord with the value from our ST-FMR measurements (Supplementary Note 9). This agreement further...
discussed the excellent efficiency of TIs in spin generation and SOT-driven magnetization switching. Moreover, the robust SOT-induced magnetization switching is also observed in devices with a Cu or NiO insertion layer between the Bi2Se3 and Py layer (Supplementary Note 11 and Supplementary Figs. 14–16). From the control measurements, we find that neither the Joule heating nor the current-induced Oersted field could lead to the observed current-induced magnetization switching (Supplementary Notes 10 and 12, Supplementary Figs. 13 and 17).

Discussion

It has been suggested that interdiffusion may occur at the interface between TI and transition metals42,43, which can play a role in the TI/FM interface despite maintaining the TSS43. The exact role of interdiffusion in the switching process, however, remains unclear, and suggests that future studies focused on the TI/FM interface are necessary.

The fundamental obstacle for high-density non-volatile applications of magnetic devices in a conventional spin torque scheme is the high critical switching current density, resulting in a large size of the current driving transistor. Utilizing the giant SOT effect in Bi2Se3, which can be grown in a wafer scale using MBE, we achieve a significantly low $J_c$ to switch a conventional 3d ferromagnet NiFe, which is widely utilized in industries, addressing an outstanding scalability issue in modern magnetic devices. Moreover, no requirement of an assistive magnetic field for our demonstrated magnetization switching scheme makes the TI/FM material systems easy to integrate into the established industrial technology for magnetic devices. Our above findings may bring this exotic newly discovered quantum matter from research activities to core ingredients in real spintronics applications.

**Methods**

**Film growth and device fabrication.** High-quality Bi2Se3 films ranging from 5 to 20 QLs are grown on Al2O3 (0001) substrates in a MBE system (MBC-1000-2C from ULVAC) with a base pressure <1.5 $\times$ 10$^{-9}$ Torr. By using two-step deposition procedure44,45. The sapphire substrates are first cleaned in acetone, isopropanol and de-ionized water, and subsequently annealed at 750 °C for 30 min in a vacuum after being transferred into the growth chamber. Elemental Bi (6N) and Se (5N) solid sources are evaporated from standard Knudsen cells under a Se/Bi flux ratio of ~20. To reduce Se vacancies in Bi2Se3, initial 2–3 QL Bi2Se3 are deposited at 150 °C, and then the substrate temperature is ramped to 250 °C at 5 °C min$^{-1}$ under Se flux for the second step growth. Our Bi2Se3 films have smooth surface with a roughness of ~0.5 nm and show a clear terrace step of ~1 nm ($\approx$1 QL). The morphology indicates the high quality of our Bi2Se3 films. After the Bi2Se3 growth, the bare Bi2Se3 films are immediately transferred into a magnetron sputtering chamber via air in the standard cleanroom environment with a well-controlled levels of low humidity and constant temperature. The transfer time was strictly controlled under 5 min before pumping down the sputtering chamber. For the ST-FMR devices, a 7-nm thick Co80Fe20B20 (CFB) is subsequently sputtered on the Bi2Se3 film with a low power of 60 W at room temperature with a base pressure of ~3 × 10$^{-9}$ Torr. Finally, the Bi2Se3/CFB bilayer is protected by the sputtered MgO (2 nm)/Al2O3 (3 nm) layer. For the ST-FMR imaging devices, the Py (6 nm)/MgO (1 nm)/SiO2 (4 nm) stacks are subsequently sputtered onto the Bi2Se3 (8 QL) films with an in-situ magnetic field along the y-axis (i.e., perpendicular to the current channel, see Fig. 5f) at room temperature with a base pressure of ~3 × 10$^{-9}$ Torr. A very low sputtering power of 40 W is used for the Py deposition. Subsequently, five 2-μm wide grooves on the Py layer are etched and backfilled with nonmagnetic metal Cu, which divide the continuous Py layer into five rectangles and make them magnetically isolated. All devices are patterned by photolithography and ion milling.

**ST-FMR measurements.** The ST-FMR signals are detected by a lock-in amplifier. The frequencies and nominal power of the rf current $j_{RF}$ are 6–9 GHz and 15 dBm, respectively. An in-plane external magnetic field (H) is applied at a fixed angle ($\theta_H$) of 35° with respect to $j_{RF}$.

**MOKE imaging measurements.** The Py magnetic easy anisotropy in the device is along the $\pm y$-axis (Fig. 5f) due to the shape anisotropy. This allows us to capture the magnetization switching after pulsed dc current is off, where there is no current-induced spurious effects in the MOKE images. For the MOKE imaging measurements, we first saturate the Py magnetization along the $+ y$ or $-y$-axis with an in-plane external magnetic field $H$, then we remove $H$ and apply a pulsed dc current (500 μs pulse width) to observe the magnetization switching using MOKE microscope.

**Data availability.** The data that support the findings of this study are available from the corresponding author on request.

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Author contributions
Y.W. and H.Y.: Conceived and designed the research. D.Z., Y.W. and K.-L.T.: Carried out thin film growth and characterization. Y.W. and D.Z.: Carried out ST-FMR sample preparation and measurements with inputs from R.M., R.R. and S.S. Y.W. and Y.Y.: Fabricated the samples for MOKE measurements. Y.W. and Y.Wu.: Performed MOKE imaging measurements. Y.Wu.: Prepared the MOKE setup. Y.W. and M.E.: Performed model the analysis. Y.W., D.Z., Y.Y., R.R. and H.Y.: Prepared the figures and manuscript. All authors discussed the data and the results. H.Y.: Supervised the project.

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