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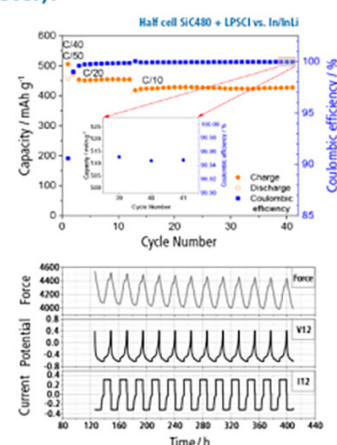
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Spin-charge conversion in wafer-scale Sb_2Te_3 topological insulatorRehab M G Ahmed^{1,2,*} , Matteo Belli³ , Emanuele Longo⁴ , Ali Shafiei¹ , Claudia Wiemer¹ , Mario Alia¹ , Graziella Tallarida¹ , Carlo Zucchetti² , Massimo Longo⁵ , Marco Fanciulli⁶ and Roberto Mantovan^{1,*} ¹ CNR-IMM, Unit of Agrate Brianza (MB), Via C. Olivetti 2, 20864 Agrate Brianza (MB), Italy² LNESS-Department of Physics, Politecnico di Milano, Piazza Leonardo da Vinci 32, Milano 20133, Italy³ CNR-IMEM Unit of Parma, Parco Area delle Scienze 37/A, 43124 Parma, Italy⁴ Institut de Ciència de Materials de Barcelona (ICMAB-CSIC), Campus UAB, Bellaterra, Catalonia 08193, Spain⁵ Department of Chemical Science and Technologies, University of Rome Tor Vergata, Via della Ricerca Scientifica 1, Roma 00133, Italy⁶ Department of Chemistry, University of Torino, Via P. Giuria 9, Torino 10125, Italy

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E-mail: rehab.ahmed@polimi.it and roberto.mantovan@cnr.it**Keywords:** topological insulator, Sb_2Te_3 , ferromagnetic resonance, interface, spin pumping, MOCVDSupplementary material for this article is available [online](#)

Abstract

Achieving wafer-scale integration of topological insulators (TIs) with silicon technology is crucial for developing future spintronic devices based on them. In particular, since spin-charge conversion (SCC) phenomena are often at the core of devices' functionalities, increasing SCC efficiency in TIs-containing spintronic devices is demanding. We report here on the room-temperature SCC performance of the Sb_2Te_3 TI as grown on a full 4-inch area Si(111) substrate by metal organic chemical vapor deposition, followed by a post-growth annealing step. In particular, SCC is probed by conducting broadband ferromagnetic resonance (FMR) and spin-pumping FMR on $\text{Sb}_2\text{Te}_3/\text{Au}/\text{Co}/\text{Au}$ heterostructures. A significant SCC was observed across the full wafer area, and the signal was predominantly attributed to the presence of topological surface states in Sb_2Te_3 . The SCC is interpreted within the Edelstein effect framework, which allows for quantifying the SCC efficiency through the inverse Edelstein effect length λ_{IEE} lying in the 0.140–0.312 nm range along the diameter of the 4-inch substrate. The observed SCC variability is connected to variations in the structure and morphology of Sb_2Te_3 across the probed wafer area and provides a technologically relevant background for the future use of TIs in spintronic devices.

1. Introduction

Topological insulators (TIs) are characterized by high spin–orbit coupling (SOC) and spin-momentum locking, which dictate the spin and momentum orientations of electrons traveling within their topologically protected surface states (TSS) [1]. These unique properties make TIs promising for several emerging spintronic device concepts, such as spin–orbit torque MRAM [2] and magnetoelectric spin–orbit devices [3]. In these architectures, converting pure spin currents into charge currents (and vice versa) is a key operating function.

The spin-charge conversion (SCC) occurring in TIs or heavy metals (HMs) is defined as the ratio between the charge current density they generate upon the injection of a pure spin current from a neighboring ferromagnetic (FM) layer. Over the last decade, various strategies to boost SCC have been explored. A common practice is to utilize FM/HM bilayers such as (Fe, Co, CoFeB, ...)/(Pt, Ta, W, ...), where the injection of spin currents from the FM layer into HMs generates an electric voltage through the so-called inverse spin Hall Effect (ISHE) [4, 5]. In TIs, SCC can additionally arise from the bulk spin Hall Effect, or more distinctively, from the spin-momentum locked TSS via the inverse Edelstein effect

Table 1. Values of various parameters extracted from the fitting of FMR and spin pumping experimental data of all the samples using equations (1)–(5).

Samples (Sb ₂ Te ₃ thickness, nm)	M_{eff} (emu/c m ³)	α	H_0 (Oe)	$g_{\text{eff,Sb}_2\text{Te}_3}^{\uparrow\downarrow}$ ($\times 10^{19} \text{ m}^{-2}$)	J_c^{2D} ($\times 10^{-6} \text{ A/m}$)	J_s^{3D} ($\times 10^5 \text{ A/m}^2$)	λ_{IEE} (nm)
n0 (0)	544 $\pm$ 2	0.0127 $\pm$ 0.0002	17 $\pm$ 6	—	6.15 $\pm$ 0.12		
n1 (35)	907 $\pm$ 3	0.0160 $\pm$ 0.0005	20 $\pm$ 6	0.950 $\pm$ 0.081	4.25 $\pm$ 0.3	3.10 $\pm$ 0.34	0.140 $\pm$ 0.014
n3 (47)	834 $\pm$ 8	0.0150 $\pm$ 0.0003	14 $\pm$ 6	0.870 $\pm$ 0.025	8.72 $\pm$ 0.6	3.37 $\pm$ 0.20	0.260 $\pm$ 0.016
n5 (40)	875 $\pm$ 6	0.0170 $\pm$ 0.0002	18 $\pm$ 7	1.130 $\pm$ 0.034	1.08 $\pm$ 1	3.46 $\pm$ 0.14	0.312 $\pm$ 0.013

(IEE) [6–8], where the presence of TSS can markedly improve the SCC figure of merit. Beyond HMs and TIs, SCC is also actively studied in Rashba interfaces [9, 10], oxide two-dimensional electron gases [11], and Weyl semimetals [12].

Typically, room-temperature SCC efficiencies λ_{IEE} , as probed by spin-pumping ferromagnetic resonance (SP-FMR) in 3D-TIs-based systems such as Bi₂Se₃, Bi₂Te₃, and Sb₂Te₃, span from 0.01 to 2.7 nm [13–17]. A comprehensive compilation of these previously reported literature values is detailed in table 1 of the recent review [18] as well as in Kumar *et al* [19]. As a comparison, HMs such as Pt exhibit spin Hall angles of 0.05–0.10, corresponding to effective λ_{IEE} values on the order of 0.1 nm when accounting for spin diffusion lengths [20]. This is the main reason λ_{IEE} values of ≥ 0.2 nm are typically considered highly competitive for room-temperature operation [21].

By making use of metal-organic chemical vapor deposition (MOCVD), some of us recently reported large room-temperature SCC and charge-to-spin conversion in heterostructures based on Sb₂Te₃ [7, 8, 22] and Sb₂Te₃/Bi₂Te₃ [23]. While sputtering is widely used for TIs, it often leads to stoichiometric deviations and higher defect densities [24]. In contrast, MOCVD offers superior control over growth conditions, enabling smooth interfaces and stoichiometric precision, which is essential for preserving TSS.

Sb₂Te₃ is a narrow-gap semiconductor that is particularly attractive due to its high thermal stability (≥ 400 °C), compatibility with industrial processes, and van der Waals (vdW) structure enabling epitaxial growth on Si (111) [25]. In particular, the SCC performance of the MOCVD-produced Sb₂Te₃ on Si (111) is competitive with that reported for other members of the 3D-TI class synthesized by other methods [26–30]. Additionally, TIs produced by MOCVD are characterized by a high potential for technology transfer, thanks to the capability of producing uniform and conformal growth over large-area substrates [31].

Within this work, we extend the investigation of SCC performance obtained in the Sb₂Te₃ TIs as MOCVD-grown on extended 4-inch Si (111) wafers [31]. In particular, we conduct room-temperature broadband FMR (BFMR) and SP-FMR on Sb₂Te₃/Au/Co/Au heterostructures. While the majority of existing literature relies on characterization at a single location or on isolated flakes, the present work uniquely provides a wafer-scale assessment of SCC. We demonstrate spatial variations in the inverse Edelstein length across the investigated positions of the 4-inch deposition area. By interpreting the SCC phenomena through the framework of the IEE effect, we report a clear variability in λ_{IEE} and correlate the observed SCC with spatially resolved variations in the structure and morphology of the Sb₂Te₃ film. Our findings highlight that interface uniformity is the critical requirement for achieving reproducible SCC in scalable TI-based systems.

2. Materials and methods

The Sb₂Te₃ thin films were grown on Si (111) substrates using an Aixtron (AIX 200/4) MOCVD system with a 4-inch rotating graphite susceptor holder. To remove the native oxide layer before deposition, the Si (111) substrate surfaces were cleaned with 5% HF, rinsed, and dried under N₂ flux, and immediately stored in the MOCVD's glove box. For the Sb₂Te₃ growth, SbCl₃ and bis(trimethylsilyl) telluride precursors were used [32]. To represent growth over the full 4-inch area, two 1.5 $\times$ 5 cm² Si(111) substrate stripes, cut with an identical orientation relative to the primary flat, were positioned radially on the MOCVD susceptor [31]. The substrates were pre-annealed at 500 °C for 60 min under N₂ (20 mbar, 11 l min^{−1}). This step was designed to desorb residual hydrogen and surface contaminants, facilitating

a well-ordered Si (111) reconstructed surface. These preparation steps are essential for minimizing interface defects and ensuring high-quality vdW epitaxy of the Sb_2Te_3 layers [32]. The deposition was then carried out at 25 °C, 15 mbar, total flow 5.075 l min^{-1} , and precursor partial pressures of 2.45×10^{-2} mbar (SbCl_3) and 3.57×10^{-4} mbar (Te precursor). This produced a low growth rate, enabling nominal film thicknesses of 40 nm after 150 min of growth. Post-growth annealing was performed at 300 °C for 15 min under an N_2 atmosphere (900 mbar, 5.575 l min^{-1}) to enhance film quality. The films were grown in a rotating-susceptor MOCVD reactor; the continuous rotation averages possible azimuthal variations in precursor flow and thermal conditions across the deposition area, making radial variations the most likely source of any residual growth non-uniformity. Additional details about the growth procedure can be found in Refs. [31, 32]

Once extracted from the MOCVD system growth chamber, each stripe was subsequently cut into smaller samples, each one identified as n1 (edge), n3 (middle), and n5 (center), as depicted in figure 1(a). Representative samples were collected from edge, middle, and center regions of the deposition area, including orthogonal radial directions, to evaluate the uniformity of the deposited films and of their SCC properties. Hereafter, these regions will be referred to by using these labels.

These samples were quickly transferred to the e-beam evaporation chamber (Edwards Auto306) to grow the Au(5 nm)/Co(5 nm)/Au(5 nm) multilayer on top. Since the e-beam evaporator used for Au/Co/Au deposition accepts pieces no larger than $1.5 \times 1 \text{ cm}^2$, a size smaller than the full stripe, this mechanical cleaving step was necessary to safely accommodate the samples within the holder. Importantly, the overgrowth of Au/Co/Au was performed simultaneously on all the n1, n3, and n5 pieces, jointly with a reference sample (n0) growth on top of Si(111). This is important to guarantee as much as possible the same chemical, structural, and magnetic quality of the Co layer in all samples, as well as to make a trustworthy comparison of their FMR and SP-FMR properties. The base pressure of the e-beam deposition chamber ranged from 5×10^{-7} to 1×10^{-6} Pa. The electron-beam current and operating pressure during the depositions were 50 mA at 4.6×10^{-6} Pa for Co and 90 mA at 7.8×10^{-6} Pa for Au.

To evaluate the crystal structure, thickness, and interface roughness of the studied samples, x-ray diffraction (XRD) and x-ray reflectivity (XRR) measurements were performed. XRD and XRR were conducted with an HRXRD IS2000 diffractometer, equipped with a Cu- $\text{K}\alpha$ radiation source, a four-circle goniometer, a commercial scintillator, and a curved 120° position-sensitive CPS-120 gas detector.

The surface morphologies and roughness of the $\text{Sb}_2\text{Te}_3/\text{Au}(5 \text{ nm})/\text{Co}(5 \text{ nm})/\text{Au}(5 \text{ nm})$ heterostructures were characterized through atomic force microscopy (AFM) measurements, conducted using a Park NX10 system in noncontact mode, exploiting sharp silicon probes with a typical tip radius of 10 nm and a resonance frequency of $\sim 320 \text{ kHz}$.

The BFMR experimental setup comprises an FMR probe with a homemade U-shaped grounded coplanar waveguide (GCPW) (see Lesne *et al* [10]) integrated with a Bruker E500 EPR spectrometer fed by an Anritsu radio frequency (RF) source in amplitude modulation (AM) mode. The employed RF ranged from 5 to 15 GHz, while the power was set to 10 dBm with 50% AM. The samples were placed face down on the GCPW in a flip-chip configuration, with the external magnetic field H_{ext} applied in the film plane and the magnetic field of the RF signal perpendicular to the DC field H_{ext} . The FMR absorption spectra were acquired by sweeping the magnetic field at several fixed frequencies.

To measure the DC voltage generated across the sample due to spin pumping, the SP-FMR analysis was performed using a nanovoltmeter connected to the FMR setup. Measurements were conducted on samples approximately 2 mm width $\times$ 3 mm length in size, obtained by cutting the n1, n3, and n5 samples defined above. A fixed microwave frequency of 9.5 GHz and an output power of 40 mW were used, while the external magnetic field was swept in-plane around the resonance condition.

Electrical connections on the sample sides were made with silver paste and copper wires. Silver paste was applied across the full sample width (2 mm) to ensure uniform potential collection, with a contact separation of 3 mm. Additionally, kapton tape was used to prevent short circuits between the sample and the GCPW [7, 33], without affecting the RF AM. All characterization experiments were repeated on samples n1, n3, and n5, in order to probe the consistency of the results in terms of SCC across the 4-inch wafer.

3. Results and discussion

3.1. Structural and morphological characterization

Sb_2Te_3 is a chalcogenide material with a rhombohedral crystalline structure belonging to the R-3 m space group [34] and characterized by vdW stacking of quintuple layers (QLs) along the c -axis. Each

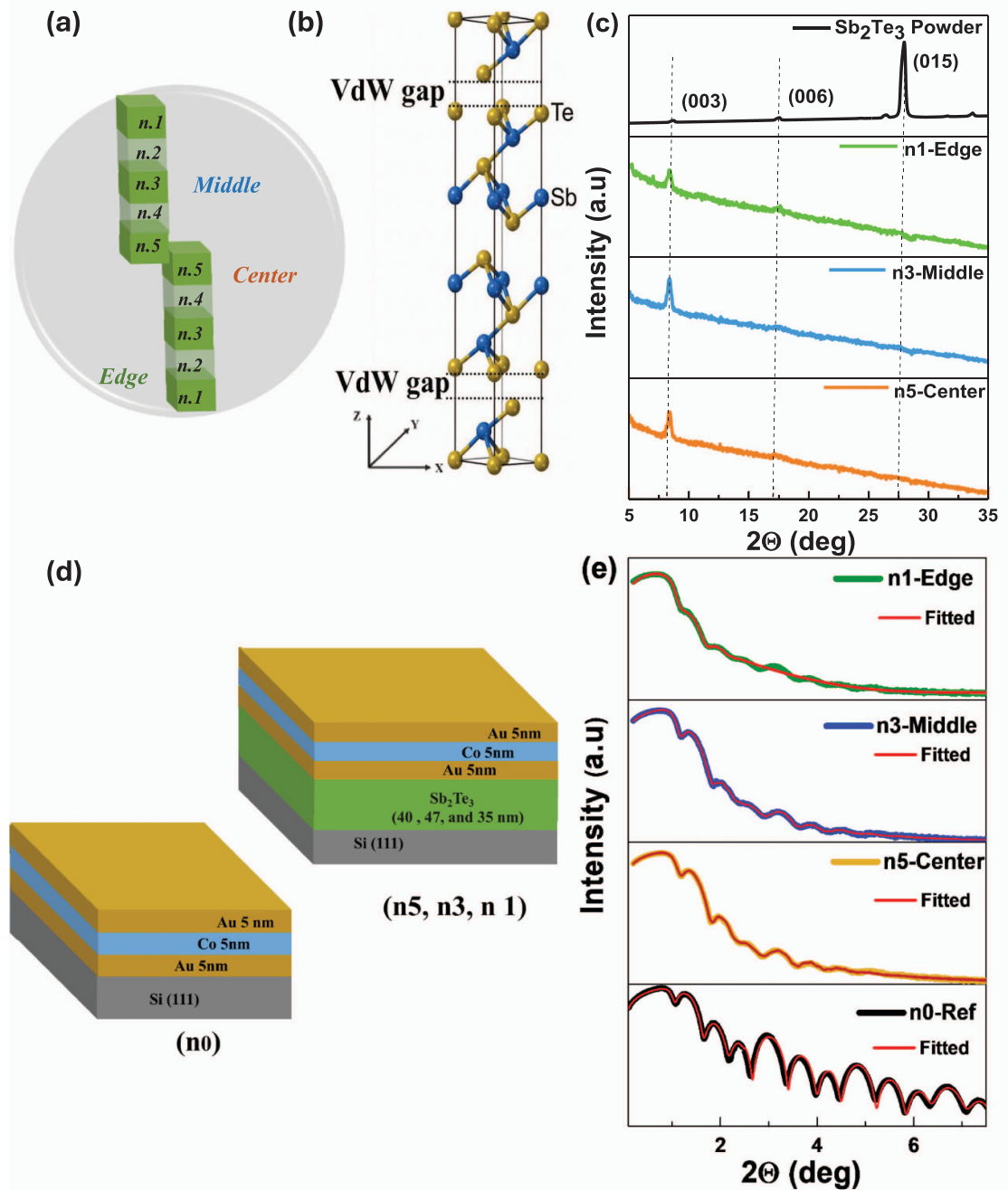


Figure 1. (a) Graphic depiction of the Si (111) substrate geometry with identified positions of the different pieces that have been cut (n1, n3, n5) and selected to conduct all the studies presented here. (b) Crystalline structure of Sb₂Te₃. (c) XRD as conducted on n1, n3, and n5 samples. (d) Schematic showing the samples structure for spin pumping measurement. (e) XRR measurement data for Sb₂Te₃/Au(5 nm)/Co(5 nm)/Au(5 nm) multilayer structures, and reference sample n0. The red line indicates the fit obtained from the data, used to accurately determine the layers' thicknesses.

QL consists of five covalently bonded atomic planes in the sequence Te–Sb–Te–Sb–Te, as displayed in figure 1(b). The overall structure of Sb₂Te₃ results in a highly preferential orientation along the [00 1] direction.

Figure 1(c) displays the XRD patterns acquired in a grazing incidence configuration of Sb₂Te₃ from the n1, n3, and n5 samples. The (003) plane is represented by the intense peak at $2\theta = 8^\circ$, which confirms that the 40-nm-thick Sb₂Te₃ film is strongly oriented along the *c*-axis. The high relative intensity of this (003) peak is a primary indicator of good texture and high crystallinity in the deposited film, as previously reported [31, 32]. The observed intensity variation in the (006) peak across the samples reflects MOCVD growth kinetics: the wafer center undergoes faster deposition, leading to a slight increase in inter-grain misorientation. The edge film (n1) benefits from precursor depletion,

which slows the growth rate and allows for enhanced adatom surface diffusion, resulting in a tighter mosaic spread. Conversely, the faster growth rates and larger accumulated volume in the central regions (n3, n5) promote greater structural misorientation as the layers accumulate. To get further insights into the Sb_2Te_3 film structural quality, ω -scan (rocking-curve) measurements were performed to assess the layers' mosaicity, reflecting a high crystalline quality (see figure S1 in the supplementary Info and Shafiei *et al* [31]).

Figure 1(d) represents the structure of the complete stack after the e-beam evaporation of the Au/Co/Au trilayer on top of Sb_2Te_3 (n1, n3, n5) and on the Sb_2Te_3 -free Si reference (n0). The thickness and interface quality were quantitatively analyzed by XRR measurements, as shown in figure 1(e). The collected XRR data were fitted using a multilayer model to extract the thicknesses and interface roughness values. The material density was fixed during the fitting, whereas the thickness and roughness were treated as free parameters. The extracted values are summarized in table S1. The Au/Co/Au layers show thicknesses close to their nominal values for all the samples, while the fitted thickness for the underlying Sb_2Te_3 layers exhibits some variation, yielding to 40 ± 1 nm, 47 ± 1 nm, and 35 ± 1 nm for the n5, n3, and n1 samples, respectively. Such a variation of the Sb_2Te_3 thickness across the wafer scale is in accordance with our previous study [31]. The thickness variation (40 nm at the center, 47 nm at the middle, 35 nm at the edge) is attributed to precursor depletion and hydrodynamics inherent to the horizontal MOCVD reactor geometry. As described in Cho and Choi [35], reactant gases are progressively depleted along the flow direction, while a stagnant boundary layer develops over the susceptor, increasing the diffusion path. These mass-transport-limited dynamics result in a non-uniform growth rate, yielding the observed radial gradient despite susceptor rotation. From XRR measurements, a change in the interface roughness has been detected in the n1, n3, and n5 samples. In particular, while the roughness of the Sb_2Te_3 interface for n5 and n3 is limited to 2.3 and 2.4 nm, respectively, the one for n1 is significantly higher, being 4.3 nm. The fitting revealed the Si substrate roughness to be uniform and very low (≈ 0.2 nm) across the entire stripe; therefore, the substrate is excluded as the source of the observed variations. The increased roughness at the wafer edge (sample n1) is thus attributed to the Sb_2Te_3 growth itself. Consequently, at the wafer edge, the morphological quality of the Sb_2Te_3 worsens, having an impact on the structural properties of the Au/Co/Au top stack and affecting the SCC performance (see later). For a detailed understanding of the chemical properties across these MOCVD-grown Sb_2Te_3 films, we refer the reader to Shafiei *et al* [31].

The surface topography of the final heterostructures, as probed by AFM (figure 2), is in agreement with XRR results. For the control sample n0, a root-mean-square (RMS) roughness of 2.0 ± 0.5 nm is measured. The RMS of n5 and n3 were found to be 2.1 ± 0.6 nm and 2.0 ± 0.3 nm, respectively, while the surface roughness of n1 shows a sharp increase up to 5.5 ± 1 nm, as expected by the much higher Sb_2Te_3 surface roughness when compared to n5 and n3 (see table S1). Although it is a relatively high value, this is primarily due to the presence of some residual isolated Au agglomerates on the sample surface. Although all samples were metallized simultaneously under identical evaporation conditions, this distinct Au agglomeration on the edge sample (n1) is dictated by the underlying Sb_2Te_3 film. Specifically, MOCVD gas flow dynamics at the wafer's edge alter local growth kinetics, creating a higher density of nucleation sites that prevent uniform Au wetting. When excluding these surface formations, the underlying film remains significantly smooth (as shown in figure 2(a)).

3.2. Dynamic magnetization and spin pumping-FMR measurement

As an example of FMR response, figure 3(a) shows the signals extracted from sample n5 for different frequencies of the RF radiation. Similar FMR curves acquired for samples n0, n1, and n3 are reported in figure S2 of the supplementary information. The FMR signal is characterized by the superposition of symmetric and asymmetric Lorentzian components, and therefore it is modeled by using a generalized Lorentzian function with the form of equation (1) [36], from which the resonance field (H_{res}) and the linewidth (ΔH) are determined for each RF frequency.

$$P_{\text{abs}} = K_1 \frac{\left(\frac{\Delta H}{2}\right)^2}{\left(\frac{\Delta H}{2}\right)^2 + (H_{\text{ext}} - H_{\text{res}})^2} + K_2 \frac{\left(\frac{\Delta H}{2}\right)(H_{\text{ext}} - H_{\text{res}})}{\left(\frac{\Delta H}{2}\right)^2 + (H_{\text{ext}} - H_{\text{res}})^2} \quad (1)$$

where K_1 , K_2 , and H_{ext} are the symmetric and asymmetric coefficients, and the applied external field, respectively [37].

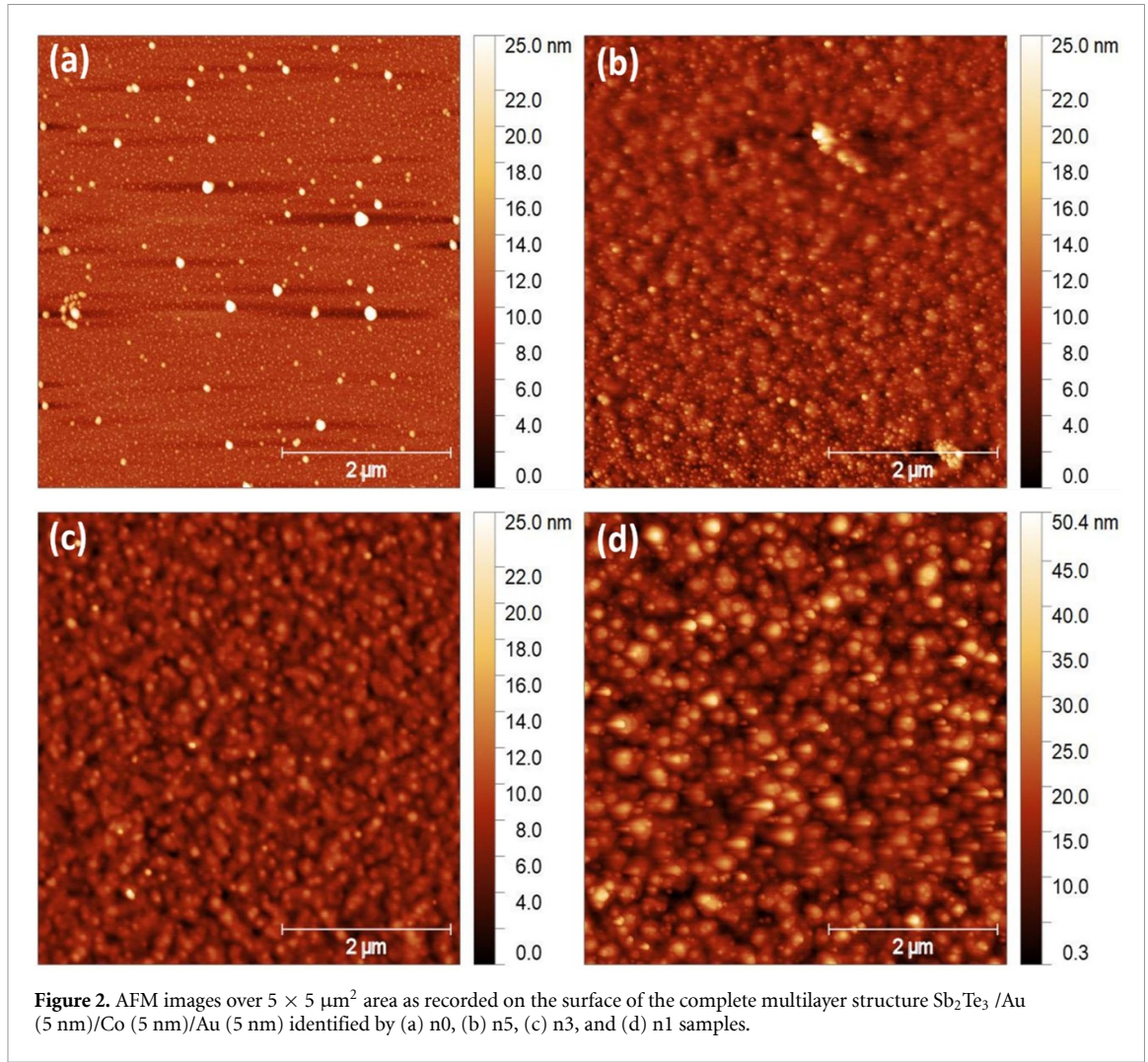


Figure 3(b) shows the plot of the resonance frequency, f_{res} vs H_{res} , for all samples. The resulting dispersion curves were fitted by using the Kittel equation (equation (2)) to determine the effective saturation magnetization M_{eff} [38]:

$$f_{\text{res}} = \frac{\gamma}{2\pi} \sqrt{H_{\text{res}}(H_{\text{res}} + 4\pi M_{\text{eff}})} \quad (2)$$

where the gyromagnetic ratio γ is defined as $\gamma = (g \times \mu_B)/\hbar$, with g being the Landé factor, μ_B the Bohr magneton, and $\hbar$ the reduced Planck constant. During the fitting process, the gyromagnetic ratio was fixed to $\gamma = 1.98 \times 10^7 \text{ Hz Oe}^{-1}$, which corresponds to a g -factor of ≈ 2.25 [39]. The obtained M_{eff} values are summarized in table 1, resulting lower than the Co bulk value (1400 emu cm^{-3}). This has been observed previously, and it is possibly due to the magnetic shape anisotropy and/or the formation of magnetic dead layers [23, 40, 41]. In our heterostructures, these dead layers are primarily located at the Co/Au interfaces. This is a well-documented effect in transition metal/noble metal systems, arising from interfacial intermixing, reduced atomic coordination, and dilute alloy formation [42]. The Au interlayer was intentionally employed to isolate the Co layer from the underlying Sb_2Te_3 , thereby preventing direct chemical interaction and stabilizing the interfacial quality.

Figure 3(c) shows the FMR linewidth as a function of f_{res} for all the samples, as interpreted and fitted within the linear relationship described by equation (3) [43].

$$\Delta H = \Delta H_0 + \frac{4\pi}{|\gamma|} \alpha f_{\text{res}} \quad (3)$$

where ΔH_0 relates to the inhomogeneity term, and α represents the damping parameter. All the samples with Sb_2Te_3 layer (n1, n3, n5) display a broader FMR linewidth compared to the reference n0, which indicates that spin pumping effects are enhanced due to Sb_2Te_3 . From the linear fit of the data

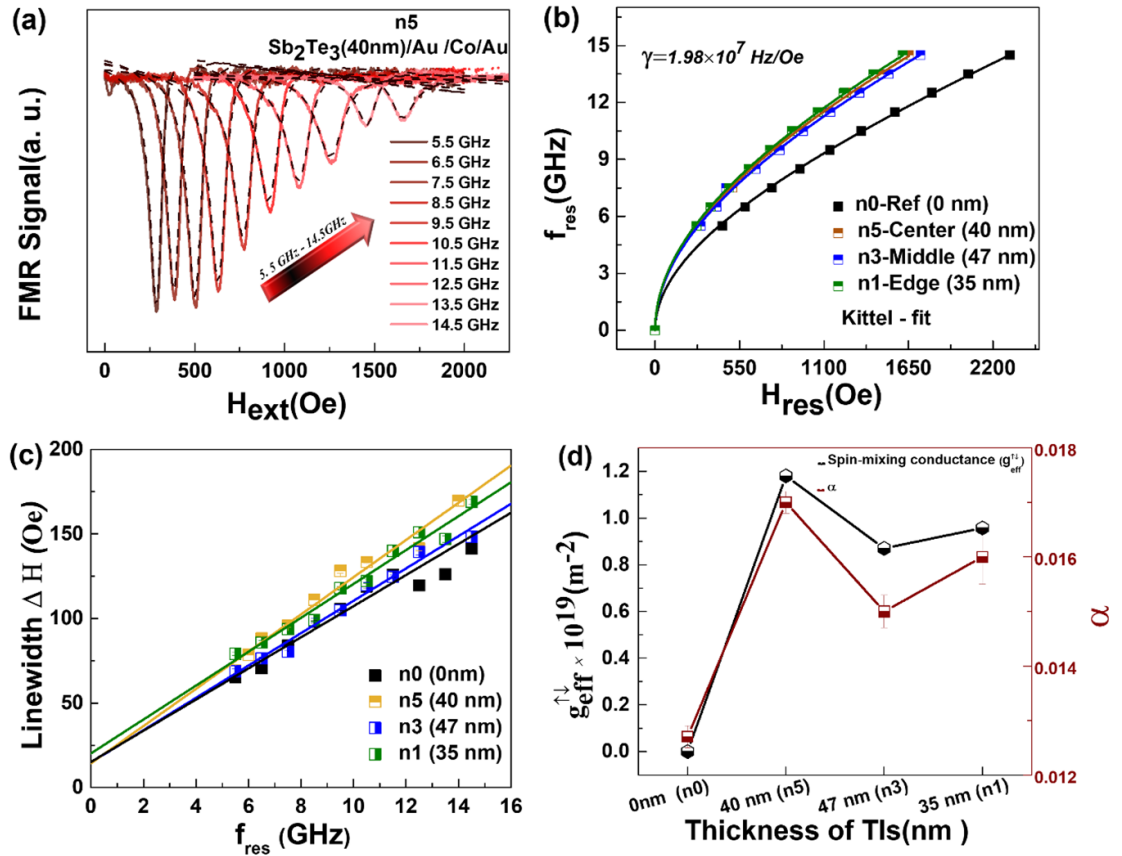


Figure 3. (a) FMR signal measured for different RF frequencies for the representative sample n5. (b) Resonant frequency f_{res} as a function of the resonance field (H_{res}), where Kittel fits are represented by solid lines; (c) FMR signal linewidth of as a function of f_{res} , where the linear fits using equation (3) are indicated with red solid lines. (d) Values of the damping constant as extracted from equation (3) and of the spin-mixing conductance ($g_{\text{eff}}^{\uparrow\downarrow}$), as calculated from equation (4) as a function of the Sb_2Te_3 thickness.

in figure 3(c), we report α values of 0.0170 ± 0.0002 , 0.0150 ± 0.0003 , and 0.0160 ± 0.0005 for n5, n3, and n1, respectively. These values are significantly larger compared to $\alpha = 0.0127 \pm 0.0002$ that we measured for n0. Figure 3(d) summarizes the trend of the obtained α values jointly with that of the spin mixing conductance $g_{\text{eff}}^{\uparrow\downarrow}$, which defines the spin transport across a non-magnetic/ferromagnetic interface, as defined by equation (4),

$$\text{Re} \left(g_{\text{eff}, \text{Sb}_2\text{Te}_3}^{\uparrow\downarrow} \right) = \frac{4\pi M_{\text{eff}} t_{\text{FM}}}{g\mu_B} \left(\alpha_{\text{Sb}_2\text{Te}_3/\text{Au}/\text{Co}/\text{Au}} - \alpha_{\text{Au}/\text{Co}/\text{Au}} \right) \quad (4)$$

where t_{FM} refers to the Co layer thickness. Interestingly, $g_{\text{eff}}^{\uparrow\downarrow}$ for n1, n3, and n5 exhibits a similar trend (and increase) as detected for α . The measured $g_{\text{eff}}^{\uparrow\downarrow}$ values for n1, n3, and n5 are of the order of $1 \times 10^{-19} \text{ m}^{-2}$, falling within the typical range previously reported for Sb_2Te_3 [7].

3.3. Electrically detected spin pumping-FMR at room temperature

The electrically detected SP-FMR analysis was carried out within the scheme presented in figure 4(a). The pure spin current J_{S}^{3D} , which is generated in the Co layers under resonance, diffuses through the neighboring Au layers finally reaching Sb_2Te_3 layer (in n1, n3, n5). When a successful SCC occurs, J_{S}^{3D} is converted into a measurable charge current J_{C}^{2D} perpendicular to the J_{S}^{3D} direction that can be detected at open circuit across the samples (see the supplementary Info for details).

First, the voltage V_{mix} was recorded through SP-FMR, which is composed by a combination of symmetric and antisymmetric contributions, as described by equation (5) [7, 44]:

$$V_{\text{mix}} = V_{\text{sy}} \frac{\Delta H^2}{\Delta H^2 + (H - H_{\text{res}})^2} + V_{\text{asy}} \frac{\Delta H(H - H_{\text{res}})}{\Delta H^2 + (H - H_{\text{res}})^2}. \quad (5)$$

The V_{sy} component refers to the SP process, while V_{asy} is due to spin rectification (SRE) effects, such as the anomalous Hall Effect and anisotropic magnetoresistance of the Co layer [45].

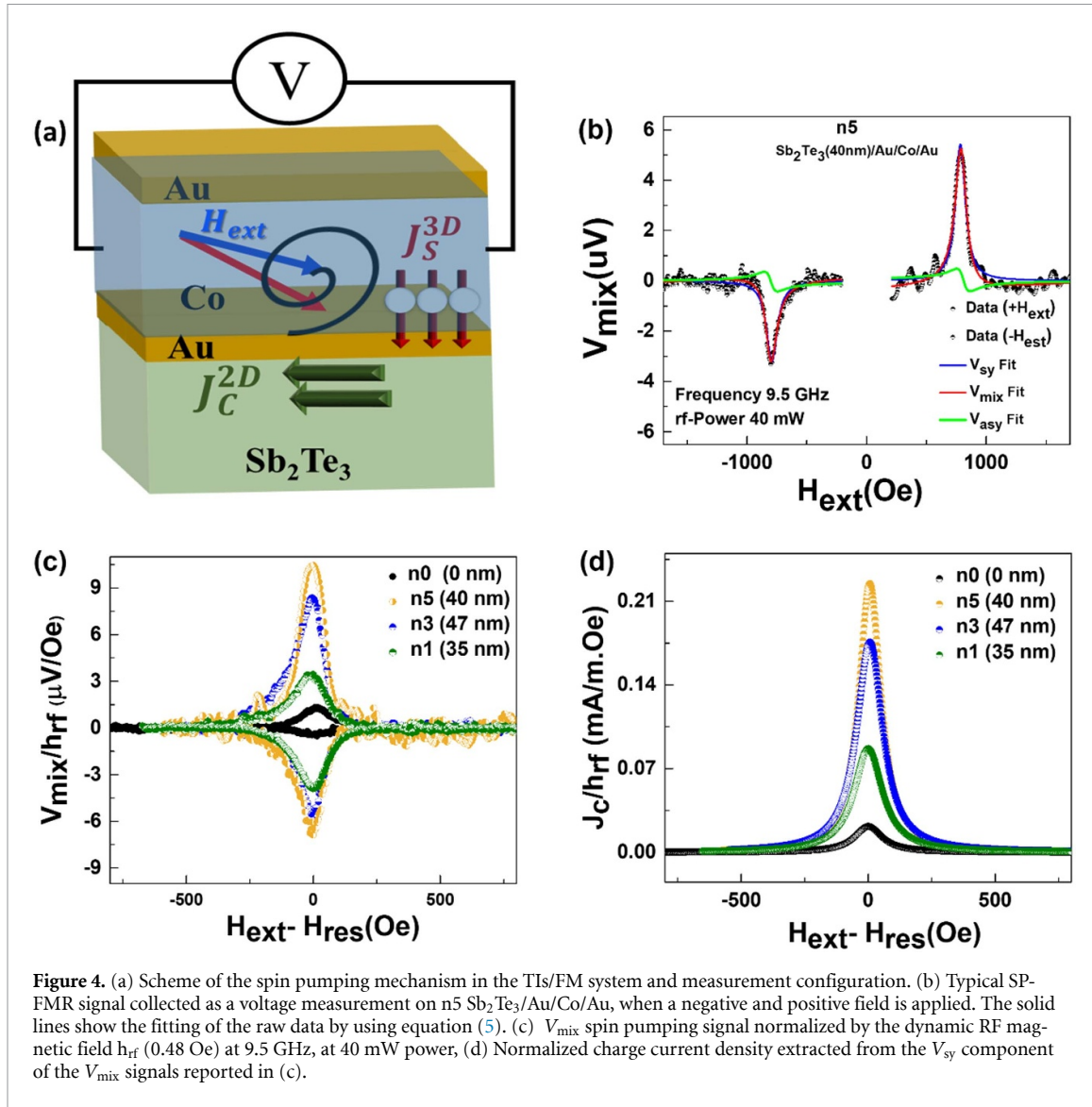


Figure 4. (a) Scheme of the spin pumping mechanism in the TIs/FM system and measurement configuration. (b) Typical SP-FMR signal collected as a voltage measurement on n5 Sb₂Te₃/Au/Co/Au, when a negative and positive field is applied. The solid lines show the fitting of the raw data by using equation (5). (c) V_{mix} spin pumping signal normalized by the dynamic RF magnetic field h_{rf} (0.48 Oe) at 9.5 GHz, at 40 mW power, (d) Normalized charge current density extracted from the V_{sy} component of the V_{mix} signals reported in (c).

Figure 4(b) shows the measured V_{mix} for n5, while the data for n1 and n3 are displayed in figure S3. For all the samples, a change in the V_{mix} sign upon the switch of the magnetic field polarity supports the spin-related origin of the signal [46, 47]. A clear symmetric contribution (blue component in figure 4(b)) was detected for all measured samples and depicted in figure 4(c) after normalizing the signal by the dynamic RF magnetic field h_{rf} . A clear enhancement of the signal for all the Sb₂Te₃-containing samples was observed when compared to n0. In order to extract the pure SP voltage contribution and get rid of potential thermal effects, the quantity V_{SP} can be defined as $V_{SP} = [V_{sy}(+H_{ext}) - V_{sy}(-H_{ext})]/2$. Finally, J_C^{2D} was calculated by using the relation $J_C^{2D} = V_{SP}/(RW)$, where R and W refer to the sheet resistance and width of the samples, respectively, and the results are summarized in figure 4(d).

The asymmetric V_{mix} in n0 originates from SRE [46]. The highly conductive Au layers sustain a large RF current, which, coupled with AMR/AHE in Co, generates a rectified voltage. These effects contribute to the antisymmetric component. The symmetric V_{sym} in n0 is not zero, and it includes a genuine ISHE contribution from the Au layers, as spin current pumped from Co flows into adjacent Au. Thus, n0 quantifies the exact background (SRE + Au ISHE). We subtract this baseline when extracting V_{sp} for the Sb₂Te₃-based heterostructures. The n0 signal is consistently ≈ 10 smaller than that from the TI samples (figure 4(d)), confirming that the dominant SCC originates from the Sb₂Te₃'s TSS. Details of spin pumping into Au are provided in Longo *et al* [7].

Furthermore, the linearity of both the V_{sy} and V_{asy} components as a function of the applied RF power (see supplementary info, figures S4(a) and (b)) demonstrates the consistency of the results with the spin-pumping mechanism and rules out significant contributions from thermal effects such as the anomalous Nernst effect [7]. As reported in figure 4(d), the comparison of the collected J_C^{2D} values

reveals a maximum 10-fold enhancement across the entire wafer area in the heterostructure where the Sb_2Te_3 layer is present, with respect to the reference n0. Despite some variations in the J_C^{2D} magnitude along the wafer, these results demonstrate the successful implementation of SCC in the MOCVD-grown Sb_2Te_3 over the full 4-inch area.

To isolate the SCC contribution of the Sb_2Te_3 layer, an Au/Co/Au stack grown simultaneously on a bare Si substrate (sample n0) serves as a baseline. Because the metallic stack grows as granular islands on the vdW Sb_2Te_3 surface but smoothly on bare Si (111), n0 is not an identical magnetic clone. These structural differences alter local anisotropies, explaining the varying resonance fields, effective magnetization, and damping. While n0 is not a perfect calibration standard for absolute magnetic dynamics, it provides the only rigorous method to quantify the non-topological background. Specifically, the Co layer's intrinsic SRE and AHE, and the Au layers' baseline ISHE. Evaluating our heterostructures against this n0 baseline indicates that the substantially enhanced symmetric voltage and 2D charge current (J_C^{2D}) are unambiguously driven by the active SCC at the Sb_2Te_3 topological surface states.

The efficiency of the SCC process is interpreted by using the IEE length λ_{IEE} , which is defined as the ratio between the generated charge current density and the injected spin current density: $\lambda_{\text{IEE}} = J_C^{2D}/J_S^{3D}$ [48]. The calculated λ_{IEE} values are summarized in table 2, where values ranging from 0.140 to 0.312 nm are reported. Notably, the λ_{IEE} for n3 (middle-region) is similar to previously reported values for a 30 nm-thick Sb_2Te_3 grown in the middle-wafer region [7], for which we expected similar chemical, structural, and morphological properties, suggesting that the SCC process is thickness-independent at least in the range from 30 to 47 nm. This reinforces the scenario of a mostly TSS-driven SCC in the presented $\text{Sb}_2\text{Te}_3/\text{Au}/\text{Co}/\text{Au}$ systems.

Crucially, the derived λ_{IEE} should be viewed as an effective length that encompasses both interfacial IEE and bulk ISHE contributions, as distinctly isolating the two remains highly challenging. The measured efficiency is also highly sensitive to the properties of the spin source itself; as reported by Longo *et al* [7] (figure S9), when Co is capped with Au and deposited on Si(111), the measured value is higher than for the identical Au/Co bilayer deposited on Sb_2Te_3 . This highlights the critical role that the substrate-induced magneto-structural properties of the Co layer play in the overall SCC efficiency. On the other hand, if the bulk conductive states of Sb_2Te_3 were the primary drivers of the conversion process, a spin pumping response should be observable in both configurations (with and without the Au buffer layer). Since this is not the case, it strongly indicates that the TSS preserved by the Au interlayer is the primary origin of the observed SCC, confirming that the inert nature of the $\text{Sb}_2\text{Te}_3/\text{Au}$ interface preserves the TSS that mediates the SCC efficiency, as recently demonstrated by some of us [62].

The Au interlayer plays a dual essential role. First, direct deposition of Co onto Sb_2Te_3 is known to cause severe interfacial intermixing and magnetic doping, which breaks time-reversal symmetry, quenching the TSS. To circumvent this issue, multiple groups have introduced non-magnetic spacer layers between the ferromagnet and the TIs. For instance, Rojas-Sánchez *et al* [15], demonstrated enhanced SCC efficiency using an Ag interlayer in Au/Fe/Ag/ α -Sn structures, while Zhang *et al* [60] employed Pt and Hf spacers in $\text{Sb}_2\text{Te}_3/\text{Pt}/\text{FeNi}/\text{Ta}$ and $\text{Sb}_2\text{Te}_3/\text{Hf}/\text{FeNi}/\text{Ta}$ heterostructures. Here, we choose Au to physically separate Co from Sb_2Te_3 and thereby preserve the intrinsic TSS. Second, Au offers a distinct advantage over many other spacer materials due to its exceptionally long spin diffusion length, which can reach up to ≈ 35 nm at RT [63]. This property is essential for ensuring that the spin current generated in the Co layer via spin pumping can efficiently traverse the spacer and reach the Sb_2Te_3 interface without significant depolarization.

The proof for the existence of TSS in our MOCVD-grown Sb_2Te_3 films was the subject of intense study in our group in recent years. While the presence of weak antilocalization (WAL) alone (as reported here in figure S5) may not be a sufficient proof, its concomitant appearance jointly with the direct observation of linearly dispersed bands by angle-resolved photoemission spectroscopy provides strong arguments supporting the presence of TSS, and this was the subject of a previous detailed study by some of us on the same material [64]. More recently, newly emerging physical features such as quantum metric have been demonstrated in our Sb_2Te_3 , a first-ever demonstration of this new phenomenon in the class of TIs materials [65]. Quantum metric is the real part of the quantum geometry tensor and measures the distance between electronic wavefunctions in momentum space [66]. It is expected to diverge close to the Dirac Point in TIs [65]. Therefore, the detection of quantum metric effects is a newcomer 'probe' to detect TSS in TIs, as demonstrated in our Sb_2Te_3 by conducting non-linear magnetotransport measurements [65].

It is also crucial to disentangle whether the observed SCC originates predominantly from the Sb_2Te_3 TSS or from a potential Au/ Sb_2Te_3 intermixed layer. In our previous work [7], this alternative alloy pathway was quantitatively evaluated using the thin-film SHE approximation, yielding an effective θ_{SHE} between 0.56 and 1.22 (based on an interfacial thickness $t \approx 1$ nm from XRR and a benchmark λ_{IEE}

Table 2. Overview of previously published values of $g_{\text{eff}}^{\uparrow\downarrow}$, R and spin-to-charge efficiency λ_{IEE} measured by FMR and SPFMR in topological materials/Ferromagnetic TIs/FM stakes.

Material	Growth method	$g_{\text{eff}}^{\uparrow\downarrow} (\times 10^{19} \text{ m}^{-2})$	λ_{IEE} (nm)	R Ω	T (K)	Ref.
$\text{Bi}_2\text{Te}_3/\text{Co}_{60}\text{Fe}_{20}\text{B}_{20}$	Sputtering	1.313	—	—	RT	[49]
$\text{BiSbTe}_{1.5}\text{Se}_{1.5}/\text{CoFeB}$	PLD	0.78	—	400	RT	[50]
$(\text{Bi}, \text{Sb})_2\text{Te}_3/\text{Ru}(0.5)/\text{CoFeB}$	MBE	1.2	0.035-0.050	600	RT	[51]
$\text{Y}_3\text{Fe}_5\text{O}_{12}/(\text{Bi}, \text{Sb})_2\text{Te}_3$	MBE	0.3-0.7	0.017-0.035	600	RT	[52]
$(\text{Bi}, \text{Sb})\text{Te}/\text{Py}$	MBE	1.0	0.075	17	RT	[53]
$\text{Bi}_2\text{Te}_3/\text{Py}$	MBE	—	0.034	692	RT	[54]
$\text{Bi}_{43}\text{Se}_{57}/\text{Co}_{20}\text{Fe}_{60}\text{B}_{20}$	Sputtering	1.8	0.32	140	RT	[55]
$\text{Co}_2\text{MnGa}/\text{Bi}_2\text{Te}_3$	Sputtering	1.243	—	10	RT	[56]
$\text{Bi}_2\text{Se}_3/\text{Bi}/\text{Fe}$	MBE	1.7	0.28	—	RT	[57]
$\text{Bi}_2\text{Se}_3/\text{CoFeB}$	MBE	3	0.36	95	RT	[58]
Sb_2Te_3 (30nm)/Au/Fe/Au(4" wafer middle)	MOCVD	0.53	0.27	16.8	RT	[9]
Sb_2Te_3 (30nm)/Au/Co/Au(4" wafer middle)	MOCVD	0.834	0.28	14	RT	[8]
$\text{Sb}_2\text{Te}_3/\text{Ru}/\text{CoFeB}/\text{Ru}$	Sputtering	3.4	0.19	75	RT	[19]
Sb_2Te_3 (15nm)/ TiO_x 1nm/CoFe 20nm	Sputtering	—	2	400	10K	[59]
Sb_2Te_3 (15nm)/Pt(1.5nm)/FeNi(5nm)	Sputtering	0.49	—	320	RT	[60]
$\text{BiSbTe}_{1.5}\text{Se}_{1.5}/\text{Cu}/\text{Ni}_{80}\text{Fe}_{20}$	PLD	1.3	1.1	825	8 K	[61]
Sb_2Te_3 (35nm)/Au/Co/Au(4-inch wafer edge)	MOCVD	0.95	0.140 ± 0.014	21	RT	This work
Sb_2Te_3 (47nm)/Au/Co/Au(4-inch wafer middle)	MOCVD	0.87	0.260 ± 0.016	22	RT	This work
n5- Sb_2Te_3 (40nm)/Au/Co/Au(4-inch wafer center)	MOCVD	1.1	0.312 ± 0.013	21	RT	This work

= 0.28–0.61 nm). Because the upper limit is unphysically large and even the lower bound of 0.56 far exceeds typical room-temperature values reported for bulk chalcogenide-based TIs, the intermixed-layer scenario is unlikely. Furthermore, this specific Au/Co/Au trilayer stack on Sb₂Te₃ has been successfully employed in our prior SCC studies [7, 8, 62]. Maintaining this well-established stack architecture enables a direct quantitative comparison with our previous benchmarks, allowing us to confidently isolate the role of the Sb₂Te₃ layer while minimizing structural variability arising from the FM/TIs interface.

It is important to note that because the heterostructures were fabricated via an *ex-situ* transfer from the MOCVD system to the e-beam evaporator, the Sb₂Te₃ films experienced a brief exposure to ambient air. This native oxidation is a critical factor that can create an interfacial barrier that can limit spin injection efficiency. However, there has been no evidence that oxidation significantly degrades the electron mobility in the surface states [64, 67]. Most likely, the surface states just migrate beneath the oxide layer without necessarily experiencing detrimental scattering. In this sense, the oxide layer protects the surface states. This is confirmed by our experimental results: the consistent spin-pumping response, robust SCC, and distinct TSS signatures in WAL measurements demonstrate that the TSS are preserved also upon surface's oxidation.

Finally, we emphasize that the overall SCC efficiency measured across the 4-inch wafer is comparable to the high values reported for other TIs systems. Table 2 summarizes the state of the art for $g_{\text{eff}}^{\uparrow\downarrow}$ and λ_{IEE} in chalcogenide-based TIs grown by different techniques (MOCVD, sputtering, PLD, and MBE), where the reported SCC efficiencies span from approximately 0.28 nm to 4 nm. Direct comparison of λ_{IEE} is most meaningful for studies based on SP-FMR under similar conditions, whereas torque-based electrical methods report different quantities (e.g. torque efficiency or θ_{SHE}), which are not directly comparable. Within the FMR SP-FMR subset, our SCC efficiencies $\lambda_{\text{IEE}} = 0.140\text{--}0.312$ nm at RT are competitive with literature values. MBE-grown (Bi, Sb)₂Te₃ and Bi₂Te₃ typically show $\lambda_{\text{IEE}} = 0.017\text{--}0.075$ nm, while higher values up to 0.32–0.36 nm are reported for Bi₂Se₃-based systems. Our central values (0.260–0.312 nm) closely match previous MOCVD-grown Sb₂Te₃ results (0.27–0.28 nm), confirming the reproducibility of our methodology. For the spin-mixing conductance, our values ($g_{\text{eff}}^{\uparrow\downarrow} = 0.87\text{--}1.1 \times 10^{19} \text{ m}^{-2}$) fall within the range reported for MBE and PLD systems, being slightly lower than some values obtained in sputtered TIs.

The unique contribution of this work, beyond reporting absolute values, is the comprehensive wafer-scale assessment of SCC. We demonstrate a significant spatial modulation in conversion efficiency across the 4-inch area, marked by a $\sim 120\%$ spread in the measured λ_{IEE} . Importantly, local variations in λ_{IEE} do not correlate with the non-monotonic film thickness profile (35–47 nm), but rather closely mirror the underlying interface quality. For instance, the wafer edge n1 exhibits the highest interface roughness and the lowest conversion length ($\lambda_{\text{IEE}} = 0.140 \pm 0.014$ nm). Conversely, the smoother center region n5 achieves our peak efficiency of ($\lambda_{\text{IEE}} = 0.312 \pm 0.013$ nm). Even with this localized landscape, SCC performance remains highly competitive across the entire wafer. This establishes MOCVD-grown Sb₂Te₃ as a viable platform for scalable TIs spintronics and definitively identifies interface uniformity as the key driver for reproducible performances.

4. Conclusion

By conducting FMR and SP-FMR measurements on Si (111)/Sb₂Te₃/Au/Co/Au heterostructures, where the Sb₂Te₃ TI was grown via MOCVD on a 4-inch wafer, we demonstrate clear and efficient SCC at room temperature. When interpreted within the IEE type of conversion, a λ_{IEE} efficiency ranging from 0.312 nm at the wafer center to 0.140 nm at the wafer periphery was found. The demonstration of relatively large SCC over the 4-inch scale indicates a high maturity level for the MOCVD production of Sb₂Te₃. These findings emphasize that precise control over interface quality is the primary requirement for achieving scalable and reproducible conversion efficiencies. Ultimately, this work provides a practical benchmark for the industrial scale-up of TIs-based spintronic devices.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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