Spin-Charge Conversion in Fe/Au/Sb₂Te₃ Heterostructures as Probed By Spin Pumping Ferromagnetic Resonance

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Large-area antimony telluride (Sb₂Te₃) thin films are grown by a metal organic chemical vapor deposition technique on 4" Si(111) substrates, and their topological character probed by magnetoconductance measurements. When interfaced with Fe thin films, broadband ferromagnetic resonance spectroscopy (BFMR) shows a clear increase of the damping parameter in Fe/Sb₂Te₃ when compared to a reference Fe layer, which may suggest the occurrence of spin pumping (SP) into Sb₂Te₃. Simultaneously, X-ray reflectivity and conversion electron Mössbauer spectroscopy evidence the development of a chemically and magnetically pure Fe/Sb₂Te₃ interface. However, by conducting SP-FMR, it is shown that no spin-to-charge conversion (S2C) occurs in Fe/Sb₂Te₃, while a clear SP signal develops by introducing a 5 nm Au interlayer between Fe and Sb₂Te₃, with a measured inverse Edelstein effect conversion efficiency of $\lambda_{\text{IEE}} = 0.27$ nm. The results shed some light on the correlation among the chemical-structural-magnetic properties of the Fe/Sb₂Te₃ interface, the broadening of the magnetic damping parameter as detected by BFMR, and the occurrence of S2C, as probed by SP-FMR.

1. Introduction

Topological insulators (TI) are gaining attention from a technological point of view due to their foreseen highly efficient capability to control adjacent magnetic media and to act as ideal spin sinks through spin-to-charge conversion (S2C) phenomena. Claritying the role played in S2C by the specific chemical bonds and/or the presence of magnetically dead-layers and/or intentionally-grown interlayers at the interface between TI and ferromagnetic (FM) layers is highly demanding. Additionally, in view of realizing a step toward technology transfer, developing fabrication methods suitable to guarantee the large-scale production of TI is necessary.

The Metal Organic Chemical Vapor Deposition (MOCVD) process has recently been successful in growing epitaxial-quality TI, such as antimony telluride (Sb₂Te₃) and bismuth telluride (Bi₂Te₃) on 4" Si(111) substrates, being characterized by the typical Van der Waals stacking order. In particular, Sb₂Te₃ layers are characterized by the expected rhombohedral crystalline structure belonging to the R-3m space group.

In the present manuscript, we show that epitaxial Sb₂Te₃ layers on Si(111) manifest improved magnetoconductance (MC) performances when compared to the “granular” Sb₂Te₃ obtained on SiO₂, and we probe S2C in heterostructures formed by Fe/Sb₂Te₃ and Fe/Au/Sb₂Te₃ stacks by making use of broadband ferromagnetic resonance spectroscopy (BFMR) and spin pumping FMR (SP-FMR).

Following the deposition of Fe layers in direct contact with Sb₂Te₃, BFMR evidences a clear increase of the damping parameter from $23 \cdot 10^{-3}$ in the reference Au/Fe/Si(111) to $41 \cdot 10^{-3}$ in Au/Fe/Sb₂Te₃, where epitaxial Sb₂Te₃ is obtained following proper thermal processing. Simultaneously, X-ray reflectivity (XRR) and conversion electron Mössbauer spectroscopy (CEMS) evidence the development of a chemically and magnetically pure interface. However, by conducting SP-FMR, we show that no S2C occurs when Fe is in direct contact with epitaxial Sb₂Te₃, while a clear SP signal emerges when a 5 nm Au interlayer is introduced to fabricate an Au(5 nm)/Fe(5 nm)/Au(5 nm)/Sb₂Te₃/Si(111) heterostructure. By interpreting the S2C conversion with the inverse Edelstein effect (IEE), a conversion efficiency of $\lambda_{\text{IEE}} = 0.27$ nm is obtained, in good agreement with that previously observed in Co/Sb₂Te₃ systems. Our results point to the conclusion that the observed S2C, as detected by SP-FMR in Fe/Au/Sb₂Te₃, originates from the topologically protected surface states (TSS) of Sb₂Te₃, which are likely detrimentally affected when depositing Fe in direct contact with Sb₂Te₃.

2. Results and Discussion

Table 1 summarizes the structure of the samples being the subject of the present work. In the AS DEP, PRE ANN, and POST
ANN heterostructures, the only difference is in the Sb$_2$Te$_3$ preparation (see Experimental Section).

The evaporated Fe and Au layers are polycrystalline, see Figure S1 in the Supporting Information. The post-growth annealing of Sb$_2$Te$_3$ triggers a reorganization of the film grains, favoring the formation of highly-oriented layers, being characterized by a surface mean roughness (as detected by atomic force microscopy) around 1 nm. A scanning electron microscopy image of the post-annealed Sb$_2$Te$_3$ is shown in Figure S2 in the Supporting Information. Moreover, the total reflection X-ray fluorescence spectrometry analysis showed mainly the presence of Sb and Te, with no additional elemental signals detected, thus demonstrating the absence of undesired contaminants in the Sb$_2$Te$_3$ layers, see Supporting Information (Figure S3, Supporting Information).

2.1. Magnetotransport in Extended Sb$_2$Te$_3$ on Si(111)

Figure 1a shows the MC (expressed by $\Delta \sigma$ in units of $e^2/h$) recorded at 10 K for the Sb$_2$Te$_3$ layers in AS DEP, PRE ANN, and POST ANN samples prior to the Fe/Au depositions. A clear non-parabolic perturbation is observed for the POST ANN sample. As a matter of fact, the MC data in all samples display a deviation from pure Lorentzian parabolic contribution, which is attributed to weak antilocalization (WAL) emerging at low temperature in the presence of highly conductive 2D states, due to intense spin-orbit coupling (SOC). Typically, the presence of WAL in TI is associated with the existence of TSS that contribute to the global conduction mechanisms.$^{[12-15]}$ In the framework of the Hikami-Larkin-Nagaoka (HLN) model,$^{[11]}$ the parameters $\alpha_{HLN}$ (being connected to the number of 2D conducting channels), and the spin coherence length ($l_\phi$), can be extracted following the fit of the MC curves, and the results are shown in Figure 1b.

Regardless of the thermal processing to which they are subjected,$^{[5]}$ all the Sb$_2$Te$_3$ layers exhibit WAL, and the extracted $\alpha_{HLN}$ ranges from 0.1 to 0.3. These values are lower than the 0.5 expected in the case of a single conductance channel (likely the surface), indicating that there is still a contribution from the bulk states. Indeed, all the layers show a consistent decrease of their sheet resistance ($R_S$) upon cooling (not shown) that, together with the high holes density ($\approx 10^{20}$ cm$^{-3}$), suggests a relevant participation of bulk states in the global conduction mechanism. The $R_S$ of Sb$_2$Te$_3$ layers lowers with thermal processing (insets in Figure 1b), due to their improved crystalline quality.$^{[5]}$ This is reflected in the marked increase of $l_\phi$ from 31 nm for as-deposited Sb$_2$Te$_3$ up to 60 nm in the post-annealed Sb$_2$Te$_3$ layer.

It is interesting to note how, even in the as-deposited Sb$_2$Te$_3$, there is a relatively high $\alpha_{HLN}$, being much higher (at same T) than that observed in as-deposited Sb$_2$Te$_3$ grown on SiO$_2$, while the $l_\phi$ is lower.$^{[11]}$ We attribute the higher $\alpha_{HLN}$ on top of Si(111) as due to a better intra-granular structural quality when compared to Sb$_2$Te$_3$ films on top of SiO$_2$, while the lower $l_\phi$ may be connected to a higher density of grain boundaries, limiting the transverse spin-coherence length.

| Sample      | Stack with layers’ nominal thickness                        |
|-------------|-------------------------------------------------------------|
| REF         | Si(111)/$^{57}$Fe(5 nm)/Au(5 nm)                            |
| AS DEP      | Si(111)/Sb$_2$Te$_3$(30 nm, As-dep)/$^{57}$Fe(5 nm)/Au(5 nm) |
| PRE ANN     | Si(111)/Sb$_2$Te$_3$(30 nm, Pre-ann)/$^{57}$Fe(5 nm)/Au(5 nm) |
| POST ANN    | Si(111)/Sb$_2$Te$_3$(30 nm, Post-ann)/$^{57}$Fe(5 nm)/Au(5 nm) |
| REF-IL      | Si(111)/Au(5 nm)/$^{57}$Fe(5 nm)/Au(5 nm)                   |
| POST ANN-IL | Si(111)/Sb$_2$Te$_3$(30 nm, Post-ann)/Au(5 nm)/$^{57}$Fe(5 nm)/Au(5 nm) |

Figure 1. Comparison of the a) MC in Sb$_2$Te$_3$ (prior to the Fe/Au deposition) in the AS DEP, PRE ANN and POST ANN samples, recorded at 10 K. The data are expressed in ($e^2/h$) unit and referred to their zero-field. b) HLN fit of the MC data, with the values of the $\alpha_{HLN}$ and $l_\phi$ parameters.
2.2. Chemical, Structural, and Magnetic Characterization of the Fe/Sb₂Te₃ Interface

In Figure 2, the chemical-structural analysis carried out by XRR on the whole set of samples listed in Table 1 is shown. Here, the collected data (black solid line) are fitted with a multilayer model and the best fit is represented by a red solid line. For the AS DEP sample, it is not possible to conduct a useful fit. This is caused by its high surface roughness of 3.9 nm. This value is comparable to the nominal 5 nm-thickness of the ⁵⁷Fe layer evaporated on top (Table 1), thus originating a non-uniform distribution of the Fe thickness, likely promoting a partial coverage of the Sb₂Te₃ surface. Indeed, the Croce-Nevot XRR model can be successfully employed for samples with a stacking order not too far from a multilayered system, where the levels of the interdiffusion or structural inhomogeneities are restrained.

Differently, clear XRR oscillations emerge in the PRE ANN and POST ANN samples (Figure 2a), showing how the improved structure and morphology of the developed epitaxial Sb₂Te₃ directly reflects in the development of a well-controlled Au/Fe/Sb₂Te₃ heterostructure. From XRR, a high chemical-structural quality is observed also for the REF-IL and POST ANN-IL samples, demonstrating the successful integration of the Au interlayer in separating the Fe layer from the Sb₂Te₃ at the bottom of the stack (Figure 2b).

The extracted XRR parameters for all the samples are summarized in Table 2. It is clear how very few differences are observed between the Au/Fe bilayers on the REF sample and the annealed Sb₂Te₃ underlayer. In the PRE ANN and POST ANN samples, we detect denser Au layers and thinner Fe layers, when compared to the REF sample. This is attributed to the different surface roughness of Si(111) and Sb₂Te₃, being respectively ≈10% and ≈50% of the total Fe thickness (Table 2). We conclude that an overall good chemical-structural quality exists for the REF, PRE ANN and POST ANN samples. Similarly to the interlayer-free samples, we detected a higher roughness roughness for the Au/Fe/Au layers deposited on top of Sb₂Te₃ with respect to Si(111) in the POST ANN-IL and REF-IL samples, respectively. On the other hand, the almost identical electronic densities for the layers in those

Table 2. XRR parameters extracted from the fit shown in Figure 2. The error bars are ±0.1 nm for the thickness (and roughness), and ±0.05 e⁻ Å⁻³ for the electronic densities.

| Layer     | Nominal Thickness [nm] | Measured Thickness [nm] | ρ(e⁻ Å⁻³) | Roughness [nm] |
|-----------|------------------------|-------------------------|----------|----------------|
| REF       | Au 5                   | 4.7                     | 5.0      | 0.48           |
|           | Fe 5                   | 5.6                     | 2.2      | 0.55           |
| Si(111)   | /                      | /                       | 0.72     | 0.4            |
| PRE ANN   | Au 5                   | 4.0                     | 4.1      | 1.6            |
|           | Fe 5                   | 3.7                     | 2.2      | 1.7            |
| Sb₂Te₃    | 30                     | 30                      | 1.2      | 1.5            |
| POST ANN  | Au 5                   | 4.2                     | 4.0      | 1.5            |
|           | Fe 5                   | 3.1                     | 2.2      | 1.7            |
| Sb₂Te₃    | 30                     | 31                      | 1.3      | 1.4            |
| REF-IL    | Au 5                   | 4.3                     | 4.5      | 0.6            |
|           | Fe 5                   | 5.8                     | 2.3      | 1.0            |
| Au 5      | 4.8                     | 4.3                    | 0.4      |                |
| Si(111)   | /                      | /                       | 0.72     | 0.4            |
| POST ANN-IL | Au 5                 | 4.4                     | 4.2      | 1.2            |
|           | Fe 5                   | 4.5                     | 2.4      | 1.2            |
| Au 5      | 5.4                     | 4.2                    | 1.3      |                |
| Sb₂Te₃    | 30                     | 28                      | 1.5      | 0.9            |

Figure 2. XRR collected data for the samples a) without and b) with Au interlayer, as listed in Table 1 (black solid line). The best fit of the curves is represented by the red solid lines and the extracted parameters are summarized in Table 2. As substantiated in the text, for the AS DEP sample a reliable XRR model could not be found due to the relatively high roughness of un-treated Sb₂Te₃.
samples, demonstrates the good reproducibility of the Au/Fe/Au deposition process on top of Sb$_2$Te$_3$ and Si(111).

Conversion-electron Mössbauer spectroscopy (CEMS) is conducted on all the samples indicated in Table 1, and the results depicted in Figure 3. In the Fe/Sb$_2$Te$_3$ structures that we previously discussed in refs. [7,10], pulsed laser deposition was employed to grow a ≈1 nm $^{57}$Fe layer in direct contact with Sb$_2$Te$_3$, then capped by a few nm of non-CEMS sensitive $^{54}$Fe. The main difference with the set of samples here under study (Table 1) is that 100% of isotopically enriched $^{57}$Fe layers are used. Consequently, CEMS probes both the top Au/Fe and bottom Fe/Sb$_2$Te$_3$ interfaces in AS DEP, PRE ANN and POST ANN, the two Au/Fe and Fe/Au interfaces in REF-IL and POST ANN-IL, and of course the whole Fe layer in all the samples. It is demanding to separate the contribution in the spectra from the top Au/Fe, in order to monitor the changes occurring at the bottom Fe/Sb$_2$Te$_3$ interface depending on the Sb$_2$Te$_3$ preparation (Table 1).

We adopt the following fitting strategy. First, the REF sample of Figure 3a is analyzed, with the aim of identifying the hyperfine parameters for the components originating at the top Au/Fe interface. These are used, and fixed (except where indicated), in the AS-DEP, PRE ANN, and POST ANN samples in Figure 3b–d, since the Au/Fe deposition is done at RT in all the samples following identical procedures and without any post-growth processing.

The CEMS-spectrum for the REF sample is dominated by a magnetically-split sextet (green SEXTET in Figure 3) that is attributed to $\alpha$-Fe, with a $B_{hf}$ slightly lower than the bulk case (32.9 T) as often observed in thin films.[17] Additionally, a small magnetic component DIST is detected (yellow in Figure 3), together with two paramagnetic doublets DOUB-1 and DOUB-2, respectively red and orange in Figure 3. All these components are still present in the PRE ANN and POST ANN samples, plus an additional paramagnetic doublet DOUB-3 (shown in Figure 3). In the REF-IL and POST ANN-IL samples, we detect $\alpha$-Fe, DIST and only the DOUB-1 paramagnetic component. All the parameters extracted from the fit of all the CEMS-spectra depicted in Figure 3, are summarized in Table 3.

The DIST in REF is due to Fe atoms partially mixing with Au at the Au/Fe interface.[18] In the AS DEP, PRE ANN and AS DEP samples, DIST now also contains the contribution from the bottom Fe/Sb$_2$Te$_3$ interface.[7,10] However, due to the relatively low spectral intensity of the DIST component, we keep a unique distribution of sextets in the fit of the AS-DEP, PRE ANN and POST ANN CEMS signal, where the contribution from the bottom Fe/Sb$_2$Te$_3$ interface becomes dominant, as indicated by the lower isomer shift when compared to the DIST in REF (Table 1), which is now very similar to that observed at the Fe/Sb$_2$Te$_3$ interfaces.[7,10] The nature of the DIST component seems not to be very much affected by the Sb$_2$Te$_3$ preparation (Figure 3 and Table 3).

Figure 3. CEMS spectra of samples a) REF and b–d) AS DEP, PRE ANN, POST ANN; e) evolution of the relative fractions of the spectral components detected in the AS DEP, PRE ANN, and POST ANN samples. f,g) CEMS spectra of samples REF-IL and POST ANN-IL containing the Au interlayer.
DOUB-1 in the REF sample could potentially originate from both the top Au/Fe or/and the bottom Fe/Si interface(s). It is well known that Fe and Si have a certain reactivity even at RT, and paramagnetic doublets arise in the CEMS spectra. Indeed, FeSi and FeAu components may have a similar quadrupole splitting around 0.7 mm s\(^{-1}\) as we detect. Being the observed isomer shift of 0.4 mm s\(^{-1}\) (Table 3) very close to that expected at the Au/Fe interface, we tend to attribute DOUB-1 mostly to the top Au/Fe interface, even if a partial contribution to the spectral intensity from FeSi cannot be ruled out. DOUB-1 is indicated with “FeAu” in Figure 3.

In the REF-IL and POST ANN-IL samples, the top Au/Fe/Au trilayer is deposited simultaneously on Si(111) and Sb\(_2\)Te\(_3\), and as expected they show basically identical CEMS signals. In both samples, there is the dominating magnetically-split sextet of \(\alpha\)-Fe (\(\approx\)81% in REF-IL and \(\approx\)77% in POST ANN-IL). The rest of the spectral intensity originates from magnetic (DIST) and paramagnetic (DOUB-1) Fe-Au interactions. Fe atoms in the DIST component are taking up to 16% and 15% of the total intensity in REF-IL and POST ANN-IL, respectively. The paramagnetic doublet DOUB-1 accounts for \(\approx\)3% and \(\approx\)8% intensity in REF-IL and POST ANN-IL respectively, likely reflecting the lower surface roughness of Si(111) when compared to Sb\(_2\)Te\(_3\), giving rise to smoother top Au/Fe/Au interfaces in REF-IL, in which Fe atoms better retain their magnetic ordering.

The AS DEP CEMS-spectrum (Figure 3b) has an overall shape much well-resembling those previously observed at the interface in the 54Fe/57Fe(1 nm)/Sb\(_2\)Te\(_3\)/SiO\(_2\)/Si heterostructures. Through the fully 57Fe-enriched Fe layer, CEMS now explores the whole 5 nm-thick Fe layer and the additional top Au/Fe interface (Table 1). Therefore, in the fit of AS DEP we first include the DOUB-1 attributed to the top Au/Fe interface, as identified in the REF sample (Table 1) and keep fixed its hyperfine parameters in the fit of all samples.

### Table 3. Hyperfine parameters for all the spectral components used to interpret the CEMS data shown in Figure 3, for the full set of samples of Table 1.

| Spectral component | \(B_{hf}\) (T) | \(\delta\) (mm s\(^{-1}\)) | \(\Delta E_Q\) (mm s\(^{-1}\)) | \(<\Gamma>\) (mm s\(^{-1}\)) | \(A_{25}\) |
|-------------------|----------------|----------------|----------------|----------------|-------|
| REF | 32.75(1) | 0.0 | 0.77(1) | 0.8(1) | 4 |
| AS DEP | 32.43(2) | 0.19(2) | 0.77(2) | 1.80(7) | 4 |
| PRE ANN | 32.43(3) | 0.40(1) | 0.8(4) | 0.55(2) | 4 |
| POST ANN | 32.43(4) | 0.41(1) | 1.0(1) | 0.56(6) | 4 |
| REF-IL | 32.70(1) | 0.0 | 1.02(2) | 0.5(6) | 4 |
| POST ANN-IL | 32.70(1) | 0.29(2) | 0.41(1) | 0.5(6) | 4 |
the Sb$_2$Te$_3$-containing samples (Table 3). It is clear how in the AS DEP sample, adding an additional doublet named DOUB-3 is necessary to justify the large central spectral intensity. This paramagnetic contribution undoubtedly originates from the bottom Fe/Sb$_2$Te$_3$ interface.

To fit the CEMS of the PRE ANN and POST ANN samples, the hyperfine parameters of all the components previously identified in REF and AS DEP are kept constant, except for the relative area intensities, the DIST's isomer shift and the $\varepsilon$-Fe's $B_{hf}$ (Table 3). In Fe/Sb$_2$Te$_3$(granular)/SiO$_2$/Si, the central paramagnetic doublet has been assigned to the formation of a FeTe-type of the compound at the Fe/Sb$_2$Te$_3$ interface,[7,10] and we attribute DOUB-3 to the formation of a “FeTe” coordination that affects 43% of the Fe atoms in the AS DEP sample. The isomer shift of the DOUB-3 component is in good accordance with those previously reported, with a larger quadrupole splitting,[7,10] thus indicating a highly disordered environment around Fe. This is most likely due to the very large surface roughness up to 3.9 nm of the as deposited Sb$_2$Te$_3$ on Si(111),[5] being basically of the order of the nominal Fe thickness. Indeed, the surface roughness of the as-deposited Sb$_2$Te$_3$ on SiO$_2$ was of the order of ~2 nm,[11] on top of which, most likely, Fe layers grow with a slightly higher conformality, thus generating Fe components with lower quadrupole splitting than in the case of AS DEP. Also, in refs. [7,10] the Fe thickness was ~10 nm, thus certainly providing a more uniform layer.

A marginal Fe oxidation is detected through DOUB-2 in AS DEP, PRE ANN, and POST ANN samples, see Figure 3e, which apparently do not evolve with the thermal history of Sb$_2$Te$_3$. No oxidation is detected in samples REF-IL and POST ANN-IL. The insets in Figure 3a–d show a pictorial view of how the diffusion is detected in samples REF-IL and POST ANN-IL. The main effect of magnetic anisotropy can also originate from the hybridization of magnetic components at the Fe/Sb$_2$Te$_3$ interface, with the CEMS data of the POST ANN (Figure 3d) basically resembling that of the REF sample (Figure 3a). We underline a remarkable difference between the evolution of the quality of the Fe/Sb$_2$Te$_3$ interface (both structurally and magnetically) following the Sb$_2$Te$_3$ growth on SiO$_2$,[10] with that now observed on Si(111). In the former case, the annealing of the Sb$_2$Te$_3$ prior to Fe deposition has been shown to have no influence on the relative abundance of the interface FeTe-type of bonding, with a constant fraction of 15–20% detected.[10] This is strikingly different from the case of Sb$_2$Te$_3$ on Si(111) here presented, where already the pre-annealing of the Si(111) before the Sb$_2$Te$_3$ deposition has an immediate effect in reducing the central paramagnetic component at the Fe/Sb$_2$Te$_3$ interface, see Figure 3c–e. We think this is mainly originated by the very efficient action of the substrate thermal treatment in reducing the Sb$_2$Te$_3$ surface roughness when grown on Si(111), as previously reported.[10]

This condition leads to an almost epitaxial growth of Sb$_2$Te$_3$ with clear van der Waals layers already in the PRE ANN case.[10] As a consequence, the subsequent Fe growth occurs more regularly, with marginal chemical mixing, finally giving an almost pure magnetically-ordered Fe layer, comparable to a reference growth directly on Si(111), see Figure 3a,c,d. Therefore, the better structural quality of Sb$_2$Te$_3$ in the PRE ANN and POST ANN samples, when compared to AS-DEP (and to granular Sb$_2$Te$_3$ on SiO$_2$), has a major role in limiting the Fe chemical interaction with Te close to the Sb$_2$Te$_3$ surface. Additionally, a role could be also played by the higher energy of the impinging Fe atoms in the case of a pulsed laser deposition,[10] when compared to e-beam evaporation (present results). To develop a theoretical model for the evolution of the Fe/Sb$_2$Te$_3$ interface upon Sb$_2$Te$_3$ annealing is of certain interest, with a special focus on the thermally induced tuning of the chemical, structural or magnetic properties. We hope our results will further boost the use of first-principles multiscale modeling in this direction.[21]

### 2.3. Broadband FMR and Spin Pumping

**FMR in Au/Fe/(Au)/Sb$_2$Te$_3$**

BFMR is employed to study the magnetization dynamics of samples REF, AS DEP, PRE ANN, and POST ANN reported in Table 1, and the acquired data are depicted in Figure 4. For the AS DEP sample, the BFMR signal is too weak to be easily revealed, likely due to the disordered morphology of the Au/Fe bilayer, as emerged from the discussion of the XRR and CEMS data (Section 2.2). Figure 4a reports the evolution of the resonant frequency ($f_{\text{res}}$) as a function of the resonant magnetic field ($H_{\text{res}}$), which is interpreted with Equation (1), for all samples.

$$f_{\text{res}} = \frac{\gamma}{2\pi} \sqrt{H_{\text{res}} (H_{\text{res}} + 4\pi M_{\text{eff}})}$$  \hspace{1cm} (1)

where $\gamma$ is the gyromagnetic ratio and $M_{\text{eff}}$ the effective magnetization. The $M_{\text{eff}}$ value can be expressed as $4\pi M_{\text{eff}} = 4\pi M_s - H_k$, where $M_s$ and $H_k$ are the saturation magnetization and the magnetic anisotropy field, respectively. The extracted $M_{\text{eff}}$ ($g$-factor) values are $586\pm38$ emu cc$^{-1}$ ($2.47\pm0.02$) and $594\pm33$ emu cc$^{-1}$ ($2.47\pm0.02$) for the PRE ANN and POST ANN, respectively, and $682\pm42$ emu cc$^{-1}$ ($2.56\pm0.02$) for the REF sample, being lower than in bulk Fe ($M_{\text{eff}} = 1700$ emu cc$^{-1}$). On the other hand, the XRR and CEMS results point to a relatively good quality of the Fe layers, therefore excluding a significant contribution from magneto-structural disorder and/or magnetic dead layers in reducing $M_{\text{eff}}$. We attribute the $H_k$ increase mainly to surface-induced anisotropy.[22–24] Even in very sharp interfaces, an additional source of magnetic anisotropy can also originate from the hybridization occurring at the Au/Fe and Fe/Sb$_2$Te$_3$ interfaces, giving rise to a local modification of the materials band structure.[22]

Figure 4b summarizes the full width at half maximum (FWHM) $\Delta H$ of the BFMR signal as a function of $f_{\text{res}}$, and the linear fit is performed using Equation (2).

$$\Delta H = \Delta H_0 + \frac{4\pi\alpha}{\gamma} f_{\text{res}}$$  \hspace{1cm} (2)

where $\alpha$ indicates the damping constant (or Gilbert parameter) and $\Delta H_0$ the so-called inhomogeneous broadening term, a quantity accounting for the magneto-structural disorder present in a FM thin film.
For the PRE ANN and POST ANN samples $\alpha = 41 \cdot 10^{-3}$, representing a 44% enhancement when compared to the REF ($\alpha = 23 \cdot 10^{-3}$). Both the Kittel dispersion (Figure 4a), and the $\alpha$ values extracted for samples PRE ANN and POST ANN (Figure 4b), show that these two contributions are magnetically very similar, which is in accordance with their almost identical chemical, structural and magnetic properties, as probed by XRR and CEMS (Section 2.2). The measured $\Delta H_0 < 30$ Oe for all samples (y-intercept in Figure 4b) is significantly lower or comparable with many reports present in the literature.\cite{29–34}

Again, this corroborates the good magneto-structural quality of the FE layer anticipated by XRR and CEMS results. Often, the increase of the FMR damping parameter $\alpha$ in FM/TI systems, when compared to FM-reference structures, is attributed to the generation of a pure spin current in the FM layer that is pumped orthogonally in the TI.\cite{29–34} When this happens, following SP, a longitudinal charge current should be generated at the FM/TI interface. In the case of an occurring pure SP mechanism, the slope of the linear fit in Figure 4b is related to the so-called spin mixing conductance $g_{\text{eff}}$, as described by Equation (3).

$$g_{\text{eff}}(\text{SbTe}) = \frac{4 \pi M T_{\text{FM}}}{\mu_B} (\alpha_{\text{FM/TI}} - \alpha_{\text{FM}})$$

where $\mu_B$ is the Bohr magneton, $T_{\text{FM}}$ the thickness of the ferromagnetic layer, $\alpha_{\text{FM/TI}}$ and $\alpha_{\text{FM}}$ are the damping constant of the samples with and without the TI layer, respectively. $g_{\text{eff}}$ is a fundamental figure of merit, which quantifies the spin current that accumulates at the FM/TI interface. Other than SP effects, an increased $\alpha$ value at FM/TI, when compared to FM, can also originate from a spin memory loss at the FM/TI interface and/or structural inhomogeneities of the FM layer.\cite{35}

The subtraction of a proper reference is considered a reliable strategy to exclude contributions not directly related to the SP process.\cite{36,38,39} Indeed, this approach allows to eliminate the SP arising at the Au/Fe interface, thus isolating the contribution solely from the bottom Fe/Sb$_2$Te$_3$ interface. The two-magnon scattering (TMS) cannot be fully accounted for with the subtraction method, and it can generate an over- or under-estimation of $g_{\text{eff}}$.\cite{34,39} On the other hand, our extracted $g_{\text{eff}}$ value for both PRE ANN and POST ANN samples is $g_{\text{eff}} = 2.2 \cdot 10^{19}$ m$^2$, being very well in accordance with those reported for similar heterostructures.\cite{25–39} and to compare with many reports present in the literature.\cite{25–39}
The enhanced $\alpha$ value is due to the large transverse 3D spin current density ($J^T$) that is generated in the Fe layer when in contact with S$_2$Te$_3$. On the other hand, this is not converted into charge current through S2C conversion, suggesting that the high SOC in S$_2$Te$_3$ may not be sufficient to generate S2C.

It is well known how the nature of TSS is ideally preserved when Ti are interfaced with non-magnetic materials (i.e., non-magnetic impurities), while the direct contact with FM layers could prevent the conservation of the TSS.

In the case of S2C mainly occurring through the TSS, it would then be clear how such a process can be detrimentally affected when interfacing Fe with S$_2$Te$_3$. To get more insight into the above scenario, we conduct BFMR and SP-FMR on POST ANN-IL and REF-IL, where an Au interlayer between Fe and S$_2$Te$_3$ is used to protect the TSS (Table 1), and the results are summarized in Figure 5.

From the analysis of the Kittel dispersion (Figure 5a), the extracted $M_{\text{eff}}$ (g-factor) values are $1183 \pm 15$ emu cc$^{-1}$ ($2.12 \pm 0.01$) and $1213 \pm 11$ emu cc$^{-1}$ ($2.01 \pm 0.01$) for the POST ANN-IL and REF-IL, respectively. When compared to the set of samples without the Au interlayer (Table 1), the $M_{\text{eff}}$ values are almost doubled. Due to the similar chemical, structural, and magnetic quality of the Fe layers in POST ANN and POST ANN-IL (Section 2.2), we attribute such a higher $M_{\text{eff}}$ to a strong reduction of the surface magnetic anisotropy due to the Au interlayer. Also the lower g-factors, when compared to the samples without Au interlayer, are in accordance with the higher crystallographic symmetry of the Fe/Au interface with respect to Fe/Sb$_2$Te$_3$, being characterized by a body center cubic (BCC)/BCC and BCC/rhombohedral symmetry, respectively.

Figure 5b shows the corresponding $\Delta H_{\text{res}}(f_{\text{res}})$ curves for the POST ANN-IL and REF-IL samples. From the linear fit with Equation (2), it emerges that both samples have an extremely low $\Delta H_{\text{res}} < 4$ Oe, an indication of a very good magneto-structural quality. POST ANN-IL shows $\alpha = 8.75 \cdot 10^{-4}$, higher than the $\alpha = 7.35 \cdot 10^{-3}$ in REF-IL. Such a measured $\Delta \alpha = 1.4 \cdot 10^{-1}$ generates a spin mixing conductance $g_{\text{eff}} = 5.3 \cdot 10^{18}$ m$^{-2}$, a value which is 76% lower than in POST ANN. Such a reduction is due to the introduction of the Au interlayers, which give rise to additional dissipative phenomena for the produced spin current (i.e., interfacial spin memory loss, scattering with impurities, thermal dissipation).

The almost perfect superposition of the Kittel curves showed in Figure 5a and the $\Delta H_{\text{res}} < 4$ Oe value extracted from Figure 5b further confirm that the Au/Fe/Au trilayers deposited on Sb$_2$Te$_3$ and Si(111) substrates, have almost the same magneto-structural properties, a fundamental prerequisite to make a reliable comparison between the two cases, to estimate the S2C due to the presence of Sb$_2$Te$_3$.

To directly evaluate the S2C efficiency in the POST ANN-IL sample, SP-FMR experiments are conducted. According to Equation (6) (see Experimental Section), the produced spin current density is $J^{ZD} = 1.15 \cdot 10^6$ A m$^{-2}$. In Figure 5c the $V_{\text{mix}}$ signal acquired for sample POS ANN-IL is reported, together with a schematic representation of the SP-FMR experiment in the inset. From the fit of the data reported in Figure 5c, $V_{\text{SP}} = 9.32$ µV is obtained, from which we calculate a generated 2D charge current $J^{ZD} = V_{\text{SP}}/RW = 0.31$ mA m$^{-2}$, where $R = 16.8$ Ohm is the resistivity value measured in four-point configuration and $W = 1.8 \cdot 10^{-3}$ m is the width of the sample. By comparing Figure 5c with Figure 4c for the POST ANN, it is clear that a large $V_{\text{SP}}$ appears thanks to the Au insertion, which clearly changes sign upon the reversal of the applied magnetic field, thus demonstrating the occurrence of S2C.

Our results demonstrate that the observed S2C cannot originate from a bulk spin Hall effect in Sb$_2$Te$_3$, otherwise it should be observed in POST ANN as well. The immediate explanation is that inserting a 5 nm Au layer between Fe and Sb$_2$Te$_3$ protects the TSS in Sb$_2$Te$_3$ (Figure 1), finally allowing to electrically detect S2C. This is in accordance with what we have previously observed in Co/Sb$_2$Te$_3$ systems, thus showing that the S2C is
not dependent on the FM nature (Fe or Co). Indeed, the successful use of Au in protecting the TSS of other TI (Bi$_2$Se$_3$) has been previously suggested,[44] with basically unaltered spin-momentum locking shown in Au/Bi$_2$Se$_3$. When compared to other metals such as Ag or Cr, Au has turned out to be more efficient in producing a chemically-inert Au/Bi$_2$Se$_3$ interface.[44–46] Our results show that this may well be the case also in the Au/Sb$_2$Te$_3$ system. Within the above scenario, the S2C can be described by the so-called Inverse Edelstein Effect (IEE),[47,48] where the IEE length $\lambda_{\text{IEE}} = \frac{J_0}{J_1}$ represents the figure of merit to quantify the S2C efficiency. For sample POST ANN-IL we obtain $\lambda_{\text{IEE}} = 0.27$ nm, which is in perfect agreement with the value obtained in Au/Co/Au/Sb$_2$Te$_3$[4] thus demonstrating the intrinsic role played by Sb$_2$Te$_3$ and its preserved TSS to origin such a large S2C. When compared to other 2nd class chalcogenide-based TI,[26,37–39,48–54] the extracted $\lambda_{\text{IEE}}$ is comparable to the highest reported so far.

Our results show that realizing a chemically and magnetically pure Fe/Sb$_2$Te$_3$ interface may not be enough to achieve a successful S2C, which is, on the other hand, achieved through the insertion of an Au interlayer to protect the TSS of Sb$_2$Te$_3$.

3. Conclusion

In summary, MC experiments show that the annealed epitaxial Sb$_2$Te$_3$ develops TSS-connected conduction. When interfaced with Fe, the enhanced chemical-structural quality of Sb$_2$Te$_3$ is connected to a remarkable suppression of intermixing occurring at the Fe/Sb$_2$Te$_3$ interface. Indeed, BFMR evidences that such an enhanced chemical, structural and magnetic quality of the Fe/Sb$_2$Te$_3$ interface gives rise to a clear broadening of the magnetic damping, which is often attributed to a successful S2C. However, our results show that this may not be the case, since no S2C is electrically detected in Fe/Sb$_2$Te$_3$ samples. On the other hand, the use of an Au interlayer at the Fe/Sb$_2$Te$_3$ interface turns out to be successful in achieving quite a large S2C, thus strongly suggesting that the observed effect mainly relates to the TSS of Sb$_2$Te$_3$, which are finally driving S2C, due to the IEE. We extract quite a large $\lambda_{\text{IEE}} = 0.27$ nm conversion factor, in very good agreement with what we have observed by using an Au/Co/Au injector on top of Sb$_2$Te$_3$.[4] This demonstrates that a successful S2C conversion can be achieved in Sb$_2$Te$_3$-based systems by employing an Au interlayer, and such a conversion does not depend on the nature of the FM electrode (Fe, Co).

4. Experimental Section

Sb$_2$Te$_3$ thin films were grown by MOCVD with an AIXTRON 200/4 setup operating with ultra-pure nitrogen carrier gas and equipped with a cold wall horizontal deposition chamber, accommodating a 4” IR-heated graphite susceptor. The Sb$_2$Te$_3$ crystalline quality improved with a cold wall horizontal deposition chamber, accommodating a 4” IR-heated graphite susceptor. The Sb$_2$Te$_3$ crystalline quality improved by appropriate thermal treatment of the substrate (at 500 °C) prior to the deposition (PRE ANN sample in Table 1), followed by in situ post-deposition annealing at 300 °C (POST ANN sample in Table 1). All the details about the Sb$_2$Te$_3$ samples preparation can be found in ref. [5]. Prior to the deposition of Fe, resistivity and MC measurements were carried out on the Sb$_2$Te$_3$ thin films by using a four points probe with the van der Pauw method, by making use of a closed-cycle cryostat.[5] Magnetic fields up to 0.8 T were applied and directed perpendicularly with respect to the sample plane (and the current). The MC measurements were conducted at a constant applied current (50 µA into the as-deposited and 100 µA in the pre- and post-annealed Sb$_2$Te$_3$), by measuring the voltage changes during the magnetic field scan. In particular, the measurements of the sheet resistance (R$_s$) as a function of the magnetic field (B) and temperature (T) were recorded on 1 x 1 cm$^2$ substrates cleaved from the grown wafers without any patterning, thus avoiding potential additional surface degradations. Following MC, the Sb$_2$Te$_3$/Si(111) pieces were transferred into an Edwards Auto306 e-beam evaporation tool, where the Au(5 nm)/$^{57}$Fe (5 nm) bilayers and Au(5 nm)/$^{57}$Fe(5 nm)/Au(5 nm) trilayers were deposited all in situ. For each evaporated element, the electronic gun deposition currents and the corresponding chamber pressures during the growths were for Au: 120 mA and 7.8 $\times$ 10$^{-6}$ Pa; for Fe: 80 mA and 4.6 $\times$ 10$^{-6}$ Pa. The Si(111) substrates used as references were cleaned with isopropyl alcohol and treated with HF prior to the evaporation processes.

![Diagram](Image)

Figure 6. Scheme of the experimental setup used for BFMR and SP-FMR experiments. For more details on the technical aspect of these measurements please refer to ref. [4].
X-Ray Reflectivity (XRR) analysis was conducted by using a Cu radiation with the $K_{\alpha}$ emission at $\lambda = 1.54$ Å and the patterns were collected with a commercial scintillator and modelled through a matrix formalism corrected by a Croce-Nevot factor. XRR measurements allowed the determination of the thickness, the roughness, and the electronic density ($\rho_s$) of each layer composing the sample. In particular, $\rho_s$ is related to the measured critical vector $Q_c$ by the relation $Q_c (\AA^{-1}) = 0.0375\sqrt{\rho_s(e^*/\AA^3)}$ (for the Cu $K_{\alpha}$ emission).\cite{33}

The isotopically enriched $^{57}$Fe layer allowed to perform CEMS analysis,\cite{30} which was carried out at RT on all samples in Table 1. CEMS was conducted in a constant-acceleration drive, with the sample mounted as an electrode in a parallel-plate avalanche detector filled with acetone gas. An $\alpha$-Fe foil at RT was used for the CEMS velocity scale calibration and all the reported isomer shifts are relative to $\alpha$-Fe. The relative area fraction in Figure 3e was obtained by assuming the same Debye-Waller factor $f = 1$ for all the components. Analysis of the CEMS data was carried out with the Vinda software package.\cite{36}

BFMR was conducted by a broadband Anritsu-MG3694C power source (1–40 GHz), connected to a home-made grounded coplanar waveguide (GCPW), where the samples were mounted in a flip-chip configuration (with the FM film close to the GCPW surface) with a 20 µm thick mylar foil stacked in between to avoid the shortening of the conduction line. The sample-GCPW system was positioned between the polar extensions of a Bruker ER-200 electromagnet maintaining its surface parallel to the external magnetic field $H_{ext}$ in the so called in-plane (IP) configuration. During the measurements, an RF current at a fixed frequency was carried toward the GCPW and the transmitted signal was directed to a rectifying diode (Wiltron, Model 70K850 NEG) which converts the RF-signal in a continuous DC-current, subsequently detected by a lock-in amplifier downward the electronic line. A scheme of the BFMR facility is shown in Figure 6. The same instrumental setup adopted for BFMR was employed to perform SP-FMR measurements. In this case, the edges of the sample were contacted with Ag paint and connected to a nanovoltmeter. For a fixed RF frequency and power, a DC-voltage was detected when the sample-GCPW system was positioned with the FM film close to the GCPW surface) with a 20 µm thick mylar foil stacked in between to avoid the shortening of the conduction line. The sample-GCPW system was positioned between the polar extensions of a Bruker ER-200 electromagnet maintaining its surface parallel to the external magnetic field $H_{ext}$ in the so called in-plane (IP) configuration. During the measurements, an RF current at a fixed frequency was carried toward the GCPW and the transmitted signal was directed to a rectifying diode (Wiltron, Model 70K850 NEG) which converts the RF-signal in a continuous DC-current, subsequently detected by a lock-in amplifier downward the electronic line. A scheme of the BFMR facility is shown in Figure 6. The same instrumental setup adopted for BFMR was employed to perform SP-FMR measurements. In this case, the edges of the sample were contacted with Ag paint and connected to a nanovoltmeter. For a fixed RF frequency and power, a DC-voltage was detected when the FMR occurred (see Figure 6).

For the SP-FMR measurements carried out in this manuscript, the RF frequency was fixed at 12.5 GHz and the RF power varied from 90 to 132 mW. According to the different RF power, the generated 3D spin current $J^{3D}$ (in units of A m$^{-2}$) was calculated by using Equation (6).\cite{37}

$$J^{3D} = \text{Re} \left[ \frac{\epsilon}{\varepsilon_0} \gamma^2 \frac{e^*}{h} \frac{2e}{\pi \alpha^2} \right] \frac{\mu_0 M_s - \sqrt{\left(\mu_0 M_s\right)^2 + 4\alpha^2}}{(\alpha^2 + \varepsilon M_s \gamma^2 + 4\alpha^2)}$$

where $h$ is the reduced Plank constant, $M_s$ the saturation magnetization, $\alpha$ the frequency of the RF-signal, $e$ the charge of the electron and $h_{RF}$ the transverse oscillating magnetic field generated by the GCPW.

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

Acknowledgements
The authors acknowledge the Horizon 2020 project SKYTOP “Skyrmion-Topological Insulator and Weyl Semimetal Technology” (FETPROACT-2018-01, n. 824123). The authors thank Dr. Lucia Nasi for providing the original AFM data previously reported.\cite{33} Dr. Martino Rimoldi, Dr. Raimondo Cecchini and Dr. Claudia Wiemer for their contribution to the development of epitaxial Sb$_2$Te$_3$,\cite{5,6} and Prof. Tobias Kamprath for fruitful discussions.

Open Access Funding provided by Consiglio Nazionale delle Ricerche within the CRUI-CARE Agreement.

Conflict of Interest
The authors declare no conflict of interest.

Data Availability Statement
Research data are not shared.

Keywords
conversion electron Mössbauer spectroscopy, ferromagnetic resonance, magnetotransport, spin-charge conversion, spin pumping, topological insulators

Received: July 15, 2021
Revised: September 3, 2021
Published online: November 5, 2021

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