

Spin-orbit torque switching of ultralarge-thickness ferrimagnetic GdFeCo

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We report on spin-orbit torque measurements in ferrimagnetic $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ films with bulk perpendicular magnetic anisotropy and thicknesses up to 30 nm. The dampinglike and fieldlike torques have been quantified using second harmonic Hall voltage detection. Both torques show an inverse linear dependence on the thickness that is indicative of the interfacial nature of the torques. The spin-Hall angle remains constant over the thickness range of 10 to 30 nm $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$. Remarkably, we find that this interfacial torque is able to switch a 30-nm-thick $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ film with a reasonably large thermal stability of $\approx 100 k_B T$.

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I. INTRODUCTION

Spin-orbit torque (SOT) devices have received attention recently due to their potential impact on next-generation memory technology [1,2]. Here SOT refers to the generation of a nonequilibrium spin accumulation at a heavy-metal (HM)/ferromagnet (FM) interface due to either the bulk spin-Hall effect in the HM [3,4] or the Rashba-Edelstein effect at the HM/FM interface [5]. This nonequilibrium spin accumulation diffuses into the ferromagnet [6], where it can reverse the magnetic order via the spin-transfer torque mechanism [7,8].

In most SOT studies as of today, transition-metal magnets, such as Co or Fe with MgO capping layers, have been used [9]. Here, the perpendicular magnetic anisotropy (PMA) originates from the hybridization effects at the ferromagnet/oxide interface [10]. Due to the inherent interfacial nature of this anisotropy, the ferromagnetic films must be grown very thin (on the order of 1 nm), and the Co/MgO interface must be clean and have the right oxygen stoichiometry. This imposes a fundamental bottleneck on the scaling of SOT devices, because it results in a reduction of thermal stability if the area of these devices is reduced.

One way to overcome this problem is to use magnetic materials with bulk PMA [11], where the effect of decreasing footprint area can be compensated by increasing the thickness of the magnet. Indeed, SOT switching has recently been investigated in HM/ferrimagnet structures with bulk PMA, where the ferrimagnets are transition-metal (TM)-rare-earth (RE) alloys [12–18].

However, a detailed study of the thickness dependence of SOT in magnets with bulk PMA is still lacking in literature. Such a study is important for two reasons: First, the thickness dependence of the torque components can shed light onto the underlying physics of the SOT itself; second, the thickness dependence of the film is relevant for memory applications due to the potential for scaling. It should be noted that previous studies on the thickness dependence of SOT in magnets with bulk PMA were limited to ultrathin magnetic films of just a few nanometers in thickness due to changes in the crystalline structure with thickness [19].

In the work presented here, we investigate the thickness dependence of SOT in ferrimagnetic $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ films. We vary the thicknesses from 10 to 30 nm and show that all films can be switched with 200- μs current pulses through adjacent HM layers. To measure the effective magnetic field due to SOT, we perform harmonic Hall measurements. The effective field is found to be inversely proportional to the ferrimagnet film thickness and saturation magnetization. In addition, field-switching experiments are performed to measure the thermal stability. It is found that the critical switching current j_c , as well as the thermal stability Δ , increase linearly with thickness, so that the switching efficiency j_c/Δ is constant.

II. SAMPLE PREPARATION

We deposited a series of $\text{Ta}(5)/\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}(t)/\text{Pt}(5)$ films (thickness in nm) on thermally oxidized silicon substrates by rf magnetron sputtering. The thickness t was varied from 10 to 30 nm in steps of 5 nm. The base pressure during

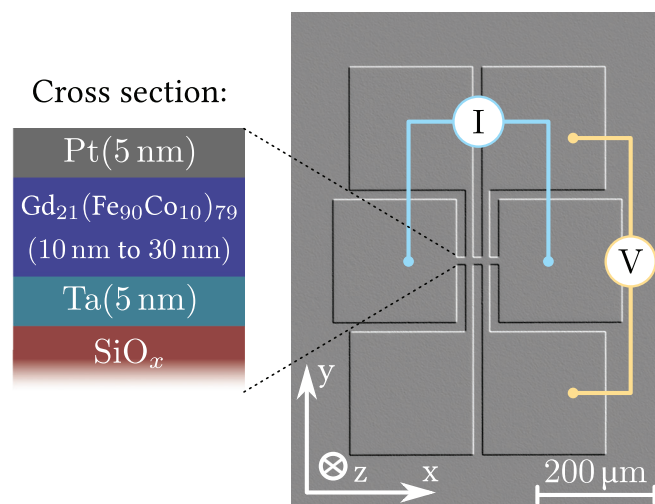


FIG. 1. Microscope image of a Hall bar device. Transition-metal-rich $\text{Ta}(5)/\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}(t)/\text{Pt}(5)$ (thickness in nm) structures are investigated. A current is applied along the x direction, and the anomalous Hall voltage is detected along the y direction. The width of the Hall bar is 20 μm .

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deposition was below 1×10^{-8} Torr. After deposition, Hall bar mesa structures were defined by optical lithography and Ar-ion milling. Figure 1 shows an optical image of a typical Hall bar device and the measurement geometry. The width of the Hall bar is 20 μm . A current is applied along the x direction to excite SOT dynamics. The response of the magnet is detected by the anomalous Hall effect (AHE), measured perpendicular to the direction of current flow, along y . Note that $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ is a transition-metal-rich alloy with $|m_{\text{TM}}| > |m_{\text{RE}}|$. The resistivity of the $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ is $\rho = 1.19 \text{ m}\Omega \cdot \text{cm}$.

III. RESULTS AND DISCUSSION

$\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ is a TM-RE alloy with antiferromagnetic coupling between the TM and the RE magnetic moments [20]. The AHE in TM-RE ferrimagnets is proportional to the out-of-plane component of the TM magnetization, while the RE magnet does not contribute significantly to the AHE [21,22]. This is because the conduction electrons in the TM are spin polarized, while the RE has no spin-split conduction band.

The AHE resistivity ρ_{AHE} as a function of thickness t is shown in Fig. 2 on the left axis. A linear trend can be observed. In addition, the planar Hall effect (PHE) resistivity ρ_{PHE} was measured. ρ_{PHE} also follows a linear trend; however, ρ_{PHE} is two orders of magnitude smaller than ρ_{AHE} . The ratio of AHE and PHE is $\xi = \rho_{\text{PHE}}/\rho_{\text{AHE}} \approx 3.2\%$. The right axis of Fig. 2 shows the saturation magnetization M_s as a function of thickness. M_s is constant for all samples with $t > 10 \text{ nm}$. However, the thinnest sample of this series with $t = 10 \text{ nm}$ has a larger value of M_s . We attribute this increase in M_s to a small shift in composition.

To characterize the SOT in our samples, we performed harmonic Hall measurements of the effective magnetic fields, following Hayashi *et al.* [23]. We started by measuring the first and second harmonic voltage responses $V_{1\omega}$ and $V_{2\omega}$

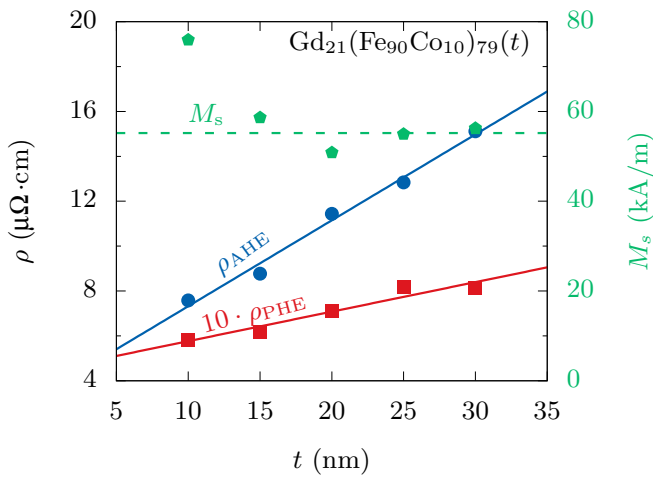


FIG. 2. The AHE resistivity ρ_{AHE} (blue, left axis) as well as the PHE resistivity ρ_{PHE} (red, left axis) depend linearly on the $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ thickness. The saturation magnetization was measured with vibrating sample magnetometry and is plotted in green on the right axis. M_s is constant for samples with thickness $t > 10 \text{ nm}$. At $t = 10 \text{ nm}$, the magnetization is enhanced.

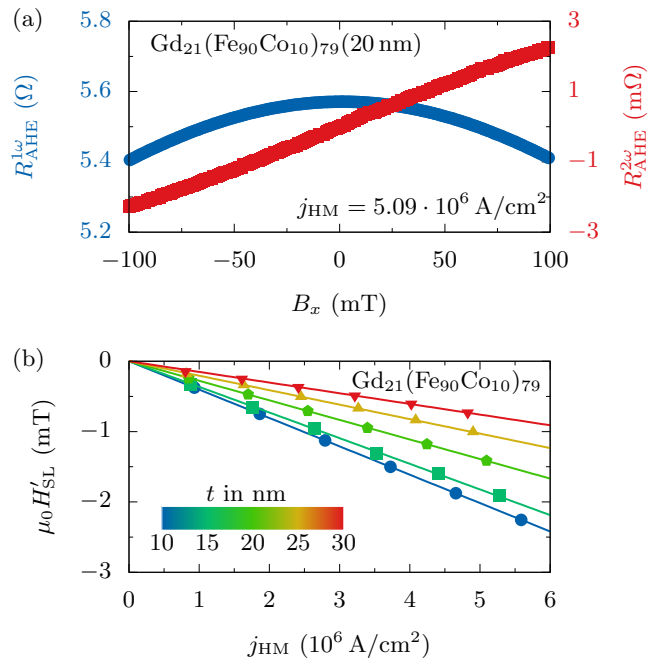


FIG. 3. Harmonic Hall measurement of the effective magnetic field induced by SOT. Panel (a) shows the first and second harmonic Hall resistance as a function of in-plane magnetic field B_x parallel to the current direction for a 20 nm thick $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ film. From this measurement the Slonczewski effective field $\mu_0 H'_{\text{SL}}$ can be extracted. Panel (b) shows that $\mu_0 H'_{\text{SL}}$ depends linearly on the current density in the heavy metal in all films under investigation. $\mu_0 H'_{\text{SL}}$ is plotted as measured and not corrected for the planar Hall effect in this figure.

to an ac current with $\omega = 1.2 \text{ kHz}$ as a function of in-plane magnetic field $B_{x,y}$. To measure the Slonczewski field H_{SL} , the in-plane magnetic field B_x was applied along the x direction, while the in-plane magnetic field B_y was applied along the y direction to measure the fieldlike field H_{FL} . For all harmonic Hall measurements discussed in the following, the samples were magnetized along the $+z$ direction, and thus $m_{\text{tot}} > 0$.

A typical harmonic Hall measurement to determine H_{SL} is shown in Fig. 3(a) for a $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ sample with $t = 20 \text{ nm}$ and $m_{\text{tot}} > 0$. The first harmonic response $R_{\text{AHE}}^{1\omega} = V_{1\omega}/I_{\text{FM}}$ (blue) follows a quadratic trend as a function of the in-plane magnetic field B_x , while the second harmonic response $R_{\text{AHE}}^{2\omega} = V_{2\omega}/I_{\text{FM}}$ (red) can be approximated by a linear function of B_x . Here I_{FM} is the current through the $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$.

To compute the effective Slonczewski field from this measurement, the slope of $R_{\text{AHE}}^{2\omega}$ is divided by the curvature of $R_{\text{AHE}}^{1\omega}$ [23]:

$$\mu_0 H'_{\text{SL,FL}} = \left(\frac{\partial R_{\text{AHE}}^{2\omega}}{\partial B_{x,y}} \right) \times \left(\frac{\partial^2 R_{\text{AHE}}^{1\omega}}{\partial B_{x,y}^2} \right)^{-1}. \quad (1)$$

Figure 3(b) shows that the effective magnetic field $\mu_0 H'_{\text{SL}}$ obtained from this analysis is proportional to the current density in the HM layer in all samples under investigation. This confirms that the effect measured here is indeed caused by SOT and not by nonlinear effects such as heating. In the

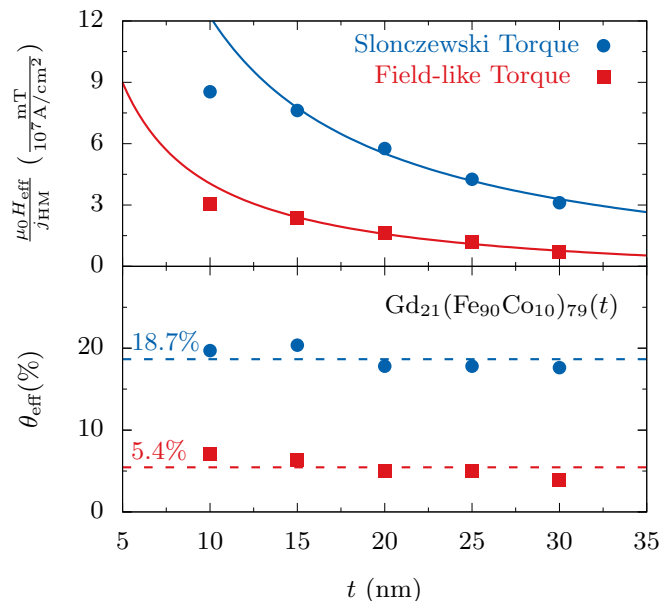


FIG. 4. The top panel shows the spin torque efficiency as a function of thickness. We find that both the dampinglike (blue dots) and the fieldlike (red squares) torque efficiencies decrease inverse linearly with thickness. The bottom panel shows the extracted spin-Hall angle θ_{SH} for the thickness series. Since the thickness t was varied but the HM layers were kept unchanged (and thus the HM/FM interface is constant), the effective spin-Hall angle does not depend on the thickness of the FM.

following, we will refer to the effective field per unit current density through the HM layers as $\zeta_{\text{SL,FL}}$.

To account for a mixing of H'_{SL} and H'_{FL} due to the presence of the PHE, the following correction was applied to the data shown in Fig. 3 [23]:

$$H_{\text{SL,FL}} = -2 \frac{H'_{\text{SL,FL}} + 2\xi H'_{\text{FL,SL}}}{1 - 4\xi^2}. \quad (2)$$

Here, ξ is the ratio of PHE resistivity and AHE resistivity: $\xi = \rho_{\text{PHE}}/\rho_{\text{AHE}}$, and the sample is magnetized upwards ($m_{\text{tot}} > 0$). The resulting fields $\zeta_{\text{SL,FL}}$ are plotted in the top panel of Fig. 4 as a function of $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ thickness. All samples with $t > 10$ nm follow a $1/t$ dependence. This is consistent with an interfacial torque of the form

$$\tau_{\text{SOT}} = \frac{\hbar}{2e} \frac{j_s}{M_s t} \mathbf{m} \times (\boldsymbol{\sigma} \times \mathbf{m}), \quad (3)$$

where e is the electric charge, \hbar the Planck's constant, and j_s the spin current. The sample with $t = 10$ nm does not follow the trend, because M_s is larger (cf. Fig. 2) and thus $H_{\text{SL,FL}}$ is smaller.

Next, we used the results for $\zeta_{\text{SL,FL}}$ to calculate the spin-Hall angle (SHA) θ_{SH} according to

$$\theta_{\text{SH}} = \frac{2|e|\hbar}{M_s t_{\text{FM}}} \frac{\mu_0 H_{\text{SL,FL}}}{j_{\text{HM}}}. \quad (4)$$

Perfect interface transparency was assumed. Figure 4 shows that the SHA θ_{SH} does not depend on t . This is expected, because the SHA only depends on the choice of SH metal and the interface. Both did not change upon increasing the

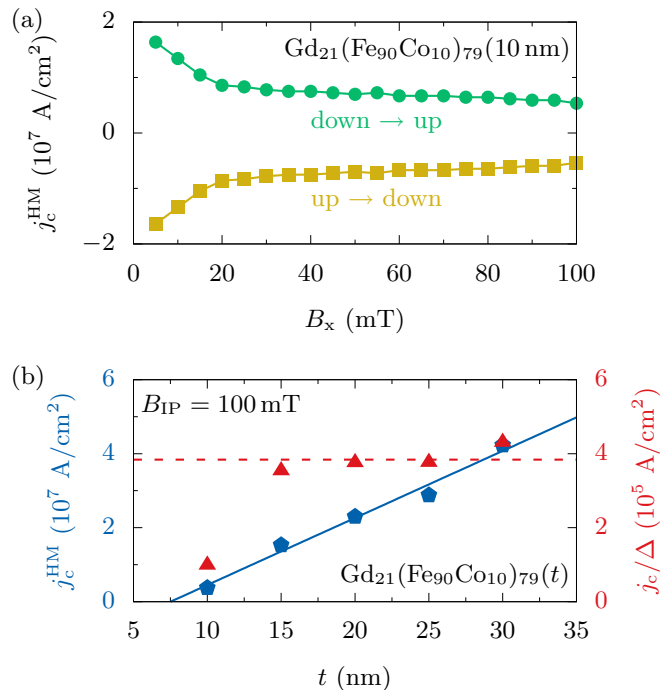


FIG. 5. Panel (a) shows the switching phase diagram for a $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ film with thickness $t = 10$ nm. Panel (b) shows that the critical switching current density depends linearly on the film thickness (blue diamonds, left axis). The ratio j_c/Δ is constant for samples with $t > 10$ nm (red triangles, right axis).

$\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ thickness. The average value for the SHA is $\theta_{\text{SH}} = 18.7\%$. This value is well in agreement with literature values in Pt/Co/Ta systems [24].

In addition to harmonic Hall measurements, we conducted SOT switching experiments to measure the scaling of the critical current density. 200- μs -long current pulses were applied through the heavy-metal layers, and the magnetic response was measured by the AHE in the $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$. In order to switch a magnetic film with PMA, an additional magnetic field B_x needs to be applied in the x direction to break the symmetry [25,26]. First, we investigated how the critical current that is needed to switch the magnet scales with B_x . A typical switching phase diagram for a $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ film with $t = 10$ nm is shown in Fig. 5(a). As expected, the critical current decreases with increasing B_x . It was further confirmed that the switching direction is reversed when we change the sign of the in-plane magnetic field [27].

Next, we show that the critical current density for SOT switching j_c at fixed $B_x = 100$ mT scales linearly with t [blue pentagons in Fig. 5(b)]. This is expected from macro spin simulations if $B_x < \mu_0 H_k$ [28], where H_k is the anisotropy field. The magnitude of j_c is similar to previous results in GdFeCo [15].

Magnetic memory applications require not only low switching current densities but also high thermal stability $\Delta = E_B/k_B T$, where E_B is the activation energy barrier. To investigate the scaling behavior of Δ with t , we performed field-switching experiments with an external magnetic field B_z applied along the magnetic easy axis. Field switching is a thermally activated process. Thus, it is expected that

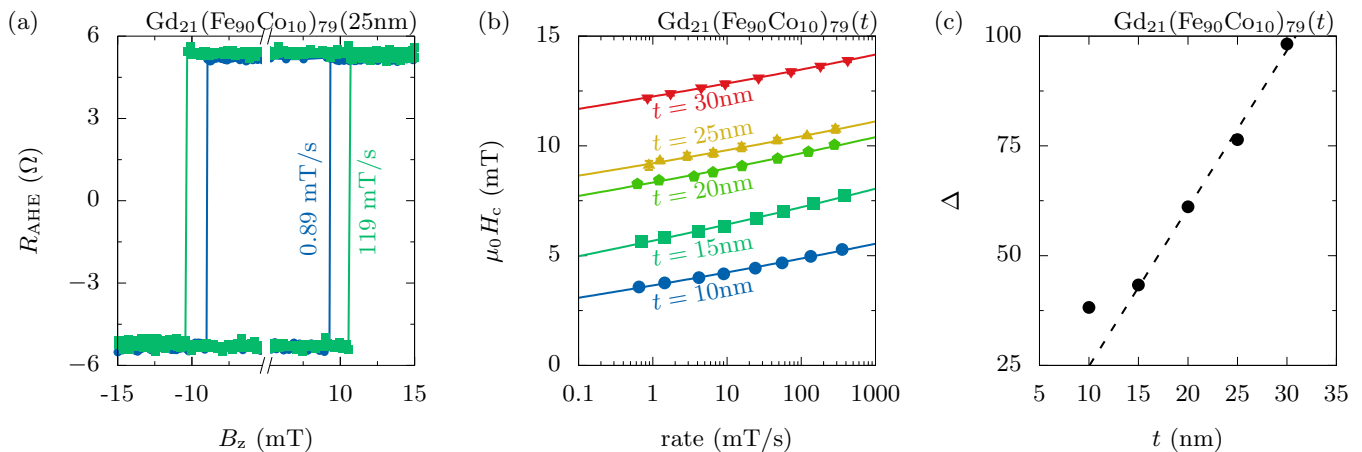


FIG. 6. Measurement of the thermal stability Δ . Panel (a) shows typical hysteresis loops for the films under investigation. B_z is applied along the easy axis of the magnet. The coercive field $\mu_0 H_c$ is increasing with the sweep rate of the magnetic field due to the thermal nature of the switching events. Panel (b) shows the dependence of $\mu_0 H_c$ on the magnetic field sweep rate. The solid lines are fits to the model by El-Hilo *et al.* [29], which is used to determine the thermal stability factor Δ . The extracted values for Δ are plotted in panel (c) as a function of $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ thickness.

the coercive field H_c depends on the rate at which the magnetic field is changed. This is shown in Fig. 6(a). At slow sweep-rates, the switching event occurs at smaller fields, as it is more likely to thermally nucleate a domain, which will then propagate in the magnet.

The dependence of H_c on the sweeping rate has been studied systematically over 3 orders of magnitude on samples with different GdFeCo thickness. The result is shown in Fig. 6(b): In a semilogarithmic plot, H_c depends approximately linearly on the sweeping rate. Note that each value for H_c is the average value from 20 consecutive field-switching measurements. To analyze this data, we used a model proposed by El-Hilo *et al.* [29]:

$$H_c = H_k \left(1 - \sqrt{\frac{1}{\Delta} \ln \left[\frac{f_0 H_k}{2\Delta} \frac{1}{r} \right]} \right). \quad (5)$$

Here, H_k is the anisotropy field, Δ the thermal stability factor, f_0 the attempt frequency, and r the sweeping rate. Figure 6(b) shows that this model (solid lines) fits the experimental data well. The extracted values for Δ are shown in Fig. 6(c). For the extraction we assume an attempt frequency of $f_0 = 1 \times 10^{10}$ Hz.

A linear trend is seen in Fig. 6(c), because the thermal stability is proportional to the magnetic volume. The sample with $t = 10$ nm does not follow the trend, which we attribute again to an increase in saturation magnetization.

One metric of switching efficiency for SOT devices is j_c/Δ . In our devices, j_c as well as Δ scale linearly with thickness. Thus, the switching efficiency does not depend on the thickness as shown in Fig. 5(b) on the right axis (red triangles).

All results discussed so far have been obtained on $\text{Gd}_x(\text{Fe}_{90}\text{Co}_{10})_{100-x}$ samples with $x = 21\%$. However, it is known that the magnetic properties of TM-RE alloys change drastically with composition [30]. In fact, previous experiments have shown that SOT exhibits a distinctive dependence on the film composition [14–17]. So far, the composition dependence has only been studied in very thin films with $t \leq 5$ nm. Further, the effective fields were estimated from

domain wall motion experiments [31]. Here, we used harmonic Hall measurements to characterize the composition dependence of the effective SOT fields in 30-nm-thick $\text{Gd}_x(\text{Fe}_{90}\text{Co}_{10})_{100-x}$ films.

Following Eq. (3), we expect the effective magnetic field to diverge at the magnetization compensation point, since M_s will vanish. This trend can clearly be seen in Fig. 7, where we plot $\mu_0 H_{\text{SL,FL}}/j_{\text{HM}}$ vs the concentration x for the damping like field, as well as for the fieldlike field.

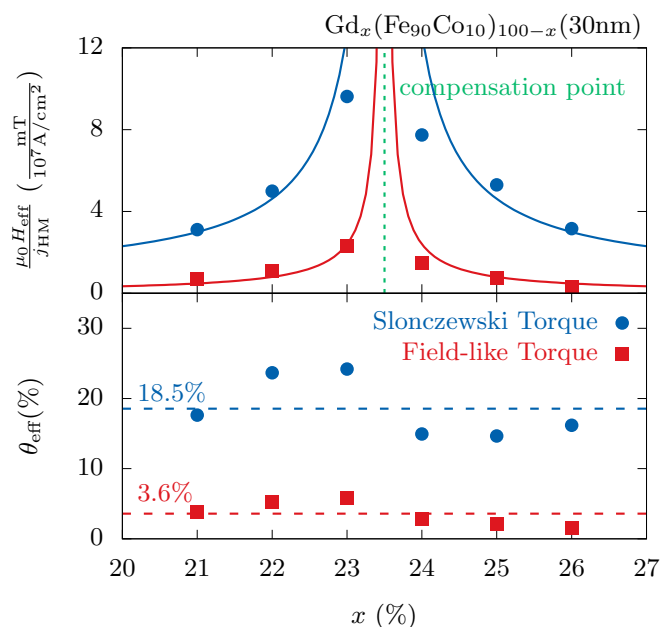


FIG. 7. (Top) The spin torque efficiency as a function of film composition x is shown for the $t = 30$ nm sample. The spin torque efficiency diverges close to the compensation point. (Bottom) The effective spin-Hall angle θ_{SH} is plotted as a function of composition x . θ_{SH} shows no distinct dependence on the composition x , since the HM/FM interface is unaltered.

Next, we calculate θ_{eff} using the values obtained for $\mu_0 H_{\text{eff}}/j_{\text{HM}}$. Figure 7 shows that θ_{eff} is approximately constant, since only the composition x but not the SH metal or the interfaces were changed. The average value for the SHA is $\theta_{\text{eff}} = 18.5\%$. This value is well in agreement with the value of θ_{eff} reported in Fig. 4.

IV. CONCLUSION

In conclusion, we showed that the effective magnetic field $H_{\text{SL,FL}}$, induced by SOT, scales as $H_{\text{SL,FL}} \propto M_s^{-1} t_{\text{FM}}^{-1}$ for all samples under investigation in this study. This indicates that the torques are interfacial in nature. θ_{SH} is found to be constant at an average value of $\theta_{\text{SH}} = 18\%$. We have also shown that $\text{Gd}_{21}(\text{Fe}_{90}\text{Co}_{10})_{79}$ as thick as 30 nm can be switched with SOT. j_c , as well as Δ , were found to scale linearly with thickness, keeping the switching efficiency j_c/Δ constant.

Thus, our work, together with torque and thermal stability measurements as a function of thickness, provides a self-consistent picture of how the SOT scales in ferrimagnet/heavy-metal heterostructures. The ability to switch large thickness ferrimagnets with high thermal stability might have important implications for future magnetic-memory applications.

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