Impact of the crystal orientation on spin-orbit torques in Fe/Pd bilayers

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Received 5 February 2020, revised 21 April 2020
Accepted for publication 30 April 2020
Published 16 June 2020

Abstract
Spin-orbit torques in ferromagnetic/non-magnetic heterostructures offer more energy-efficient means to realize spin-logic devices; however, their strengths are determined by the heterostructure interface. This work examines the impact of crystal orientation on the spin-orbit torque efficiency in different Fe/Pd bilayer systems. Results from spin torque ferromagnetic resonance measurements evidence that the damping-like torque efficiency is higher in epitaxial than in polycrystalline bilayer structures while the field-like torque is negligible in all bilayer structures. The strength of the damping-like torque decreases with deterioration of the bilayer epitaxial quality. The present finding provides fresh insight for the enhancement of spin-orbit torques in magnetic heterostructures.

Keywords: epitaxy, spin orbit torques, spin torque ferromagnetic resonance, magnetic heterostructure, Gilbert damping

(Some figures may appear in colour only in the online journal)

1. Introduction

Spin torque based spin-logic devices have been identified as prime candidates for beyond Moore technologies because of their good scaling behavior and significant potential to operate at ultra-low power and high speed. Spin torques can be classified as spin transfer torques (STTs) and spin-orbit torques (SOTs) [1, 2]. STT based spin logics have limitations due to the stochastic nature of the STT-switching and a high switching critical current density, which makes these devices unsuitable for ultrafast operation at the subnanosecond regime and beyond. However, SOT based logic operation relying on torques generated by the net spin accumulation at the magnetic/nonmagnetic interface has the potential to overcome these limitations. A basic building block for spin-orbit torque devices is the ferromagnetic (FM) and nonmagnetic (NM) bilayer with its FM/NM interface. However, SOT based logic operation relies on torques generated by the net spin accumulation at the magnetic/nonmagnetic interface has the potential to overcome these limitations. A basic building block for spin-orbit torque devices is the ferromagnetic (FM) and nonmagnetic (NM) bilayer with its FM/NM interface. The net spin accumulation at the interface can be generated by several methods, e.g. through the Rashba–Edelstine effect [3], spin Hall effect (SHE) [4], Seebeck effect [5], etc. An elegant way to generate a SOT is provided by the SHE, which may exhibit both bulk and interfacial contributions [3]. The SHE generated spin current exerts a torque on the magnetization of the free magnetic layer via a spin angular momentum transfer mechanism between the different orbitals; known as SOT. This SOT can be damping- or field-like depending on the FM/NM interface [2, 4]. Zhang et al [6] have reported the importance of the interfacial transparency for the strength of the SOT in polycrystalline FM/Pt (FM = Py, Co, Co1−xNix) heterostructures. They concluded that the electronic band matching is important for the enhancement of the SOT in bilayers. Later, Zhou et al [7] reported the effect of a collinear antiferromagnetic state on the SOT efficiency in Py/L10−IrMn bilayers. It was concluded that the coherent collinear magnetic order at the interface induces a large SOT efficiency. Lee et al [8] have reported significant enhancement of the SOT in the polycrystalline Pt/CoFeB system by the interface modification involving a Ti layer, which was attributed to enhancement of the transparency and reduction of magnetic proximity at the interface. Furthermore, it
has also been evidenced that the dominant spin relaxation or SHE mechanism in metallic NM layers follows the Elliot–Yafet mechanism, where the spin Hall angle is proportional to the resistivity of the NM layer [2, 4, 9, 10]. However, it is also reported that in epitaxial NM Ta and Pt layers, the Dyakonov–Perel mechanism originating from interfacial spin-orbit interaction (SOI) provides the dominating spin relaxation mechanism [11–13]. In case of the Dyakonov–Perel mechanism, the spin Hall angle is independent of the NM layer resistivity. Therefore, it is intriguing to understand the SHE and associated SOT efficiency in epitaxial magnetic heterostructures that are devoid of or exhibit very small interfacial Rashba SOI.

The interfacial spin torques are very sensitive to the crystallographic structure and therefore to the orbital ordering at the interface. Hence, in this study we have examined the effect of crystallographically ordered and disordered interfaces on the SOT efficiency in the Fe/Pd system. We have performed spin torque ferromagnetic resonance (ST-FMR) measurements at different applied dc currents to extract the change of the effective damping of the Fe/Pd bilayer in three different epitaxial heterostructures and one polycrystalline heterostructure. The observed critical current density at which the effective damping reverses its sign is significantly lower in the epitaxial Fe/Pd structures than in the polycrystalline Fe/Pd structure, which is evidence of a dominating of anti-damping spin torque in the epitaxial Fe/Pd structures.

2. Experimental

Fe/Pd bilayers were deposited on Si/SiO₂ and MgO substrates using an ultra-high vacuum sputtering system. To remove surface contamination and to improve surface quality of the substrates, the substrates were heat treated at 620 °C for 2 h prior to deposition. The polycrystalline Fe/Pd bilayer was grown at room temperature on a Si(100)/SiO₂ substrate, while the different epitaxial Fe/Pd/Pd bilayers were grown on MgO(100) substrates at different substrate temperatures: 238 °C, 288 °C, and 338 °C. The thickness of the Fe and Pd layers were kept constant of 5 nm. The polycrystalline film is hereafter referred to as poly-Fe/Pd, while the epitaxial Fe/Pd films are named as epi-Fe/Pd_238, epi-Fe/Pd_288 and epi-Fe/Pd_338, respectively.

The crystallographic orientation and mosaicity of the epitaxial bilayers were examined by x-ray diffraction (XRD) measurements. The individual layer thickness and interface roughness were obtained using x-ray reflectivity (XRR) measurements. The scans covered the 2θ range 0°–6°, and the XRR results were analysed using the Panalytical X’pert Reflectivity software package with a combined genetic and segmented algorithm model. Magnetic characterization of the polycrystalline and epitaxial bilayers were performed by performing in-plane angle dependent X-band cavity FMR measurements. The X-band cavity FMR measurements were performed at the 9.8 GHz frequency while sweeping the in-plane applied magnetic field during the measurements.

SOT measurements were performed by employing the ST-FMR technique [14]. ST-FMR spectra were recorded on 10 µm (width, w) × 100 µm (length, l) patterned Fe(5 nm)/Pd(5 nm) structures along the easy axis of magnetization. The measured resistances of poly-Fe/Pd, epi-Fe/Pd_238, epi-Fe/Pd_288, and epi-Fe/Pd_338 patterned structures are \( R_{\text{poly-Fe/Pd}} = 179 \), \( R_{\text{epi-Fe/Pd}_238} = 117 \), \( R_{\text{epi-Fe/Pd}_288} = 115 \) and \( R_{\text{epi-Fe/Pd}_338} = 109 \), respectively. Resistivity values of Fe and Pd epitaxial layers are nearly equal, and found to be 11.7 µΩ·cm, 11.5 µΩ·cm, and 10.9 µΩ·cm for 238 °C, 288 °C, and 338 °C grown Fe and Pd layers, respectively. In ST-FMR measurements, the microwave current \( I_m \) was injected along the sample length; see figure 1 of [14]. The ST-FMR spectra were recorded by scanning the in-plane magnetic field at 45° (\( \phi \)) with respect to the direction of \( I_m \) at different constant frequencies ranging from 9 to 16 GHz and for different applied dc current densities \( J_m \). These measurements used an internal amplitude modulation technique, where a 50% amplitude modulation of the microwave signal at 211 Hz was used for lock-in detection. The applied microwave power was kept constant at 10 dBm during measurements (for more details of the experimental setup, see [14]). The non-uniformity of the microwave power inside the patterned bar is negligible for the studied structures.

3. Results and discussion

Figure 1(a) shows the XRD pattern of the epi-Fe/Pd_238 film, which clearly evidences the Fe(200) and Pd(200) crystallographic orientations. It is known that the Fe(100) crystallographic plane rotates 45° in-plane to match the plane of
MgO(100) while the Pd(100) plane rotates 45° with respect to Fe(100) to grow epitaxially on the Fe(100) film. To confirm the epitaxial growth, ω scans were performed across the Fe(200) and Pd(200) orientations as shown in figures 1(b) and (c), respectively. The presence of a sharp single peak in respective rocking curve (ω scan) confirms the epitaxial orientation of the Fe(200)/Pd(200) bilayers. The ω scan determined value of the full width at half maximum (FWHM) for the Fe(200) orientation for the epi-Fe/Pd₂₃₈, epi-Fe/Pd₂₈₈ and epi-Fe/Pd₃₃₈ films are 2.8°, 3.0°, and 3.5°, respectively, while the value of 0.15° for the Pd(200) peak is nearly constant for all films. Figure 2 shows XRR patterns along with simulated lines for the bilayer films. The XRR determined values for individual layer thickness and roughness are presented in table 1. The interface roughness of polycrystalline and epi-Fe/Pd bilayers is in the range of 0.39–0.67 nm, confirming that the different bilayer samples exhibit sharp interfaces. There is no significant difference in interface roughness comparing polycrystalline and epi-Fe/Pd bilayers.

To extract information on the magnetic anisotropy and effective magnetization, in the in-plane angle dependent cavity FMR data were analysed by using the following equation [15]

$$ f = \frac{\gamma H_0}{2\pi} \left\{ \frac{H_c \cos(\phi_H - \phi_M) + H_u \cos 4(\phi_M - \phi_C) + H_u \cos 2(\phi_M - \phi_u)}{H_c \cos(\phi_H - \phi_M) + M_{eff} + H_u (3 + \cos 4(\phi_M - \phi_C)) + H_u \cos^2(\phi_M - \phi_u)} \right\}^{1/2}, $$

(1)

where $H_r$ is the resonance field, $f$ is the cavity microwave frequency and $\gamma$ is the gyromagnetic ratio. The in-plane directions of the applied magnetic field, magnetization, cubic and uniaxial anisotropies are given by $\phi_H$, $\phi_M$, $\phi_C$ and $\phi_u$, respectively, with respect to the (100) direction of the Si and MgO substrates. $H_c = \frac{4K_c}{M_s}$ and $H_u = \frac{2K_u}{M_s}$ correspond to the cubic and uniaxial anisotropy fields, with $K_c$ and $K_u$ being the cubic and uniaxial magnetic anisotropy constants, respectively, and $M_s$ is the saturation magnetization. $M_{eff} = M_s - H_u^2$ is the effective magnetization, where $H_u^2$ is the perpendicular anisotropy field of the film. Here $M_{eff}$, $H_c$ and $H_u$ are used as fitting parameters.

![Figure 2](image-url)

**Figure 2.** XRR spectra of the poly-Fe/Pd, epi-Fe/Pd₂₃₈, epi-Fe/Pd₂₈₈ and epi-Fe/Pd₃₃₈ films. Symbols are experimental data and red solid lines are fits.

The angle dependent FMR data were fitted using a derivative of the Lorentzian function to extract the angle dependent resonance field values. Figure 3 presents $H_r$ versus $\phi_H$ data together with fits according to equation (1). The $H_r$ versus $\phi_H$ behavior of the epitaxial films clearly evidences a four-fold in-plane anisotropy in all epitaxial bilayers, while the polycrystalline film exhibits a two-fold uniaxial in-plane magnetic anisotropy. Using $\frac{\gamma}{2\pi} = 28.03$ GHz/T, the fitting determined $M_{eff}$ values of all the epitaxial bilayers lie in the range 1.99–2.03 T, while the value is ~1.91 T for the polycrystalline bilayer. The in-plane cubic anisotropy field values of the epitaxial bilayers are nearly constant and lie in the range 42.6–45.2 mT, while the uniaxial anisotropy field values lie in the range of 0.06–0.82 mT. Conversely, the in-plane uniaxial anisotropy field value is 0.2 mT for polycrystalline bilayer. Nearly equal effective magnetization values for the epitaxial and polycrystalline films confirm comparable quality of the polycrystalline and epitaxial films. The anisotropy field and effective magnetization values are presented in table 1.

Figure 4(a) shows a schematic of the SOT acting on the Fe layer applying a constant dc current density $J_c$ in the Pd layer. The applied $J_c$ creates a spin accumulation at the Fe/Pd interface that acts as damping-like (DL) and field-like (FL) torques on the Fe layer magnetization $M$ [16, 17]. In ST-FMR measurements, the microwave current $I_{rf}$ in the NM layer generates an Oersted field ($H_{Oer}$) and a transverse spin current density ($J_s$) via the SHE [2, 4, 18]. Conversely, in a case of interfacial SOI and symmetry breaking at the FM/NM interface, and Rashba–Edelstine effective field is generated whose direction will be opposite to that of the Oersted field [2, 4]. The $I_{rf}$ excited temporal variation of the magnetization vector induces a time varying resistance due to anisotropic magnetoresistance of the FM layer. The varying resistance mixing with $I_{rf}$ yields a dc voltage output. At resonance, the torques due to the Oersted field and the transverse spin current contribute with anti-symmetric and symmetric profiles, respectively, to
the FMR line-shape, while the Rashba–Edelstein effect generated field-like torque contributes with an anti-symmetric profile [2–4, 16–18].

The SOT induced time evolution of the magnetization vector $\vec{m} = \vec{M}/M_s$ is governed by the Landau–Lifshitz–Gilbert–Slonczewski (LLGS) equation of motion, expressed as [6, 19]

$$\frac{d\vec{m}}{dt} = -\gamma \left( \frac{1}{1 + \alpha^2} \right) \left( \vec{m} \times \mu_0 \vec{H}_{\text{eff}} - \gamma \alpha \left( \frac{1}{1 + \alpha^2} \right) \left( \vec{m} \times \dot{\vec{m}} \times \mu_0 \vec{H}_{\text{eff}} \right) \right) + \frac{\gamma \mu_0 H_s}{(1 + \alpha^2) 2 \sigma M_{\text{eff}}} \left( \vec{m} \times \dot{\vec{m}} \times \vec{m} \right) - \frac{\gamma}{(1 + \alpha^2)} \left( \vec{m} \times \mu_0 \vec{H}_{\text{eff}} \right),$$

where $\alpha$ is the (Gilbert) damping parameter, $\sigma$ is the direction of the injected spin moment, $t_{\text{fe}}$ is the thickness of the ferromagnet and $H_s = H_{\text{Fe}} + H_{\text{FL}}$; $H_{\text{Fe}}$ and $H_{\text{FL}}$ are the Oersted and Rashba–Edelstein fields, respectively. The SOTs acting on the Fe magnetization are the DL torque due to the transverse spin current; $\frac{\gamma \mu_0 H_s}{(1 + \alpha^2) 2 \sigma M_{\text{eff}}} \left( \vec{m} \times \dot{\vec{m}} \times \vec{m} \right)$ and the FL torque due to the Oersted and Rashba–Edelstein fields; $- \frac{\gamma}{(1 + \alpha^2)} \left( \vec{m} \times \mu_0 \vec{H}_{\text{eff}} \right)$.

The recorded ST-FMR spectra of the poly-Fe/Pd and epi-Fe/Pd/238 bilayers at 11 GHz is shown in figure 4(b). The recorded voltage signal, which exhibits symmetric and antisymmetric Lorentzian weight factors, is larger in epitaxial Fe/Pd structures compared to the polycrystalline Fe/Pd structure. The ST-FMR spectrum is fitted using the expression [6, 18]

$$V_{\text{mix}} = V_S \left( \frac{\Delta H}{2} \right)^2 + (H - H_s) + V_A \left( \frac{\Delta H}{2} \right)^2 + (H - H_r)^2,$$

where $V_S$ and $V_A$ are the amplitudes of the symmetric and anti-symmetric components, respectively, of $V_{\text{mix}}$. $\Delta H$ and $H_r$ are the FWHM linewidth and resonance field, respectively. $V_S$ is proportional to the in-plane DL effective torque and $V_A$ is proportional to the out-of-plane effective torque due to the Oersted field and Rashba–Edelstein field-like torques; for more details see [17–19]. The fitting determined values of $\mu_0 \Delta H$ vs frequency ($f$) and $f$ vs $H_s$ for the poly-Fe/Pd and epi-Fe/Pd/238 structures at different applied dc currents $I_{\text{dc}}$ are presented in figures 4(c) and (d), and in figures 4(e) and (f), respectively. Absence of an $I_{\text{dc}}$-dependence of the $f$ vs $H_s$ profiles implies negligible/undetectable FL torque in all studied Fe/Pd bilayers. The $\mu_0 \Delta H$ vs $f$ data at different applied $I_{\text{dc}}$ were fitted to determine the effective damping $\alpha_{\text{eff}}$ using the expression $\mu_0 \Delta H = \frac{\alpha_{\text{eff}}}{15}\sigma f + \mu_0 \Delta H_0$, where $\Delta H_0$ is the frequency independent contribution. The $f$ vs $\mu_0 H_s$ data were fitted using the in-plane Kittel equation along hard axis, yielding $\mu_0 M_{\text{eff}} \approx 1.70(1) T$ and $\approx 2.05(1) T$ for poly-Fe/Pd and epi-Fe/Pd bilayers, respectively. The observed relative low value of $\mu_0 M_{\text{eff}}$ for polycrystalline film might be due to its pattern structure misalignment from magnetic hard axis. However, from angle dependent FMR measurements it is clear that $\mu_0 M_{\text{eff}}$ values for polycrystalline and epitaxial films are nearly equal. From the line-shape parameters, the average value of the bulk spin torque efficiency $\theta_{\text{eff}}$ is evaluated using the standard line-shape analysis method [6, 14, 18] found to be 0.066(3), 0.064(4), 0.068(6), and 0.030(4) for epi-Fe/Pd/238, epi-Fe/Pd/288, epi-Fe/Pd/338, and poly-Fe/Pd. However, this method assumes a perfectly transparent interface and that symmetric component is purely from a damping-like torque. This method disregards possible contributions from spin pumping due to the inverse spin Hall effect and interface spin losses, therefore, can yield erroneous values for the SOT efficiency [14].

The $\alpha_{\text{eff}}$ ($I_{\text{dc}} = 0$) values determined for positive field scan are $3.8(1) \times 10^{-3}$, $4.0(2) \times 10^{-3}$, $3.4(1) \times 10^{-3}$, and $8.0(1) \times 10^{-3}$ for epi-Fe/Pd/238, epi-Fe/Pd/288, epi-Fe/Pd/338, and poly-Fe/Pd.
epi-Fe/Pd_338, and poly-Fe/Pd, respectively. The $\alpha_{\text{eff}}(I_{dc})$ values determined for positive and negative field scans are presented in figure 5 for all Fe/Pd structures. The negative and positive slopes of the linear behavior of the effective damping with increasing dc current density for positive and negative field scans, respectively, clearly confirm the SOT modulation of the damping which is devoid of other extrinsic effects.

The $I_{dc}$ dependent changes of the effective Gilbert damping $\alpha_{\text{eff}}(I_{dc})$ is expressed as $\alpha_{\text{eff}}(I_{dc}) - \alpha_{\text{eff}}(I_{dc} = 0) = \frac{\sin \varphi}{(H_{r}+0.5M_{s}\sin \theta)(H_{r}+0.5M_{s}\sin \theta)} I_{c} \theta_{c}$, where $I_{c} = \frac{J_{c} R_{\text{eff}}}{A_{\text{Pd}} R_{\text{Pd}}}$ is the current density in the Pd layer, $A_{\text{Pd}}$ ($= 4.5 \times 10^{-14} \text{m}^2$) is the Pd layer cross sectional area, and $\theta_{c}$ is the effective charge-to-spin conversion efficiency [20]. In all of the Fe/Pd layers 50% of the applied dc current passes through the Pd layer. The $J_{c}$ dependent percentage changes of the effective damping, defined as $\alpha_{\text{eff}}(I_{dc}) - \alpha_{\text{eff}}(I_{dc} = 0)$, for poly-Fe/Pd, epi-Fe/Pd_238, epi-Fe/Pd_288 and epi-Fe/Pd_338 are 1.95%, 7.05%, 6.90% and 6.69%, respectively, at $J_{c} = 3 \times 10^{11} \text{A/m}^2$.

In Fe/Pd structures, dominating SOT is anti-damping in nature, which is acting against magnetic damping of the Fe layer. Therefore, effective damping of the magnetic layer changes with an increase in applied current density. The anti-damping torque strength increases with an increase in applied current density, and at certain critical current density anti-damping torque balances the magnetic damping of the Fe layer means magnetization of Fe layer can now precess without any losses. This is known as auto-oscillation of magnetization precession of the magnetic layer. The critical current density ($J_{c,\text{crit}}$) or damping-like torque efficiency at which the effective damping becomes zero or reverses sign in respective Fe/Pd structures is determined by using the applied dc current dependent changes of the effective damping values. The slope determined $J_{c,\text{crit}}$ values are $\pm 1.54 \times 10^{13} \text{A/m}^2$, $\pm 4.25 \times 10^{12} \text{A/m}^2$, $\pm 4.34 \times 10^{12} \text{A/m}^2$, and $\pm 4.48 \times 10^{12} \text{A/m}^2$ for epi-Fe/Pd, epi-Fe/Pd_238, epi-Fe/Pd_288, epi-Fe/Pd_338, respectively.

The $\alpha_{\text{eff}}(I_{dc} = 0)$ value for poly-Fe/Pd is higher than the values for the epi-Fe/Pd bilayers, while the $J_{c}$ dependent percentage change of the effective damping is lower in poly-Fe/Pd. The damping-like torque strength decreases for the epitaxial Fe/Pd bilayers grown at higher temperatures, which appears to be linked with the increase of mosaicity in the high temperatures grown films. These results infer an impact of the crystallographic orientation on the $J_{c}$ dependent effective damping modulation.

Very recently, SOTs have been studied in perpendicularly magnetized epitaxial and polycrystalline Co/Pt bilayers by using first and second harmonic Hall resistance measurements [21]. In that study, the damping-like torque in epitaxial Co/Pt was found to be 1.3 times smaller than the torque in polycrystalline Co/Pt. These results are in contrast to our results for in-plane magnetized Fe/Pd bilayers. This discrepancy results from the large spin–orbit coupling (SOC) of the Pt layer (5d group element) and that for the epitaxial bilayer one expects an abrupt potential change at the epitaxial interface, and therefore a strong interface spin loss [22]. In our study the Pd layers (4d group element) exhibit a smaller SOC strength. According to the XRR results there is no significant difference in interface roughness between the Fe/Pd bilayers. Therefore, one expects an abrupt potential change in all Fe/Pd bilayers; hence, interface spin loss cannot explain the different $\alpha_{\text{eff}}(I_{dc})$ results among the Fe/Pd bilayers. According to theoretical studies [23], the Rashba–Edelstein effect can contribute to both damping-like and field-like torques. It is known that Rashba–Edelstein effect scale with SOC strength [23], therefore both damping-like and field-like torques also scale with SOC strength. In a low SOC strength Pd layer, the presence of Rashba–Edelstein effect induced torques at the Pd interface is very small compared to other reported high SOC
layers interface with the magnetic layer as Rashba torques are undetected within our experimental method measurement limit.

In the Elliot–Yafet relaxation mechanism each momentum-scat-tering event has a certain probability P of being a spin-scattering event which means spin flip scattering time $\tau_{sf}$ is proportional to momentum-scattering time $\tau$. It provides the dominant contribution to the SHE in inversion symmetry preserved metals and it can be modulated by crystal structure and crystallographic orientation [24]. However, for Pd being a cubic system the SHE is expected to exhibit weak crystallographic orientation dependence [24]. The Pd layer thickness is less than its spin diffusion length (8 nm) [22], therefore the effect of the epitaxial interface on the damping-like torque efficiency cannot be ignored. According to Elliot–Yafet mechanism bulk spin Hall angle, hence SHE induced SOTs, increases monotonously with increase in resistivity. Here it is important to note that the interface spin Hall angle depends on the sharpness of interface, which can be significantly larger than bulk spin Hall angle value [25].

Resistivity of the films as well as damping-like torque strength is decreasing with increasing growth temperature. Therefore, in the studied epitaxial Fe/Pd heterostructures a dominating contribution to damping-like torque is the Elliot–Yafet mechanism induced interfacial and bulk spin Hall effect. The Dyakonov–Perel mechanism is observed in epitaxial 5d Ta and Pt films, which exhibit strong interfacial SOI. In our studied Fe/Pd heterstructures we have not observed any interfacial Rashba coupling, which indicates negligible interfacial SOI. However, a small contribution of Dyakonov–Perel mechanism in our studied Fe/Pd structures cannot be ignored as it was reported by C. T boon et al [26] at Py/Pd interface to explain the Pd thickness-dependent enhancement of damping, where effective damping is estimated by performing conventional spin pumping measurements. In the case of weak SOC such as Pd, both the spin-orbit field and the spin mixing of states at the interface are reduced proportionally to the SOC strength. The Dyakonov–Perel mechanism can be expected at small thicknesses (less than spin diffusion length) due to interfacial scattering that does break translational symmetry [23, 26, 27].

The Dyakonov–Perel mechanism is also proposed to describe the interface in metal heterostructures [27]. Therefore, detailed theoretical and experimental understanding is still required to unveil Dyakonov–Perel mechanism at epitaxial metallic interfaces with weak interfacial SOI, alike in Fe/Pd bilayers.

4. Conclusion

In conclusion, we have examined the impact of the crystallographic orientation in Fe/Pd bilayer structures on the spin angular momentum transfer across the interface by performing ST-FMR measurements. The effective damping value of the polycrystalline Fe/Pd bilayer is approximately two times larger than the values of the epitaxial Fe/Pd bilayers. The studied Fe/Pd bilayers exhibit only damping-like SOT and its strength is significantly higher in the epitaxial heterostructures as compared to the polycrystalline structure. These results provide directions for the realization of energy efficient spin-logic operations, or more specifically nano-oscillators for neuromorphic computing by utilization epitaxial magnetic heterostructures.

Acknowledgments

This work is supported by the Swedish Research Council (VR), Grant No. 2017-03799.

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References

[1] Ralph D C and Stiles M D 2008 *J. Magn. Magn. Mater.* 320 1190–216
[2] Manchon A, Železný J, Miron I M, Jungwirth T, Sinova J, Thiaville A, Garelo K and Gambardella P 2019 *Rev. Mod. Phys.* 91 035004
[3] Rojas Sánchez J C, Vila L, Desfonds G, Gambarelli S, Attané J P, De Teresa J M, Magén C and Fert A 2013 *Nat. Commun.* 4 2944
[4] Sinova J, Valenzuela S O, Wunderlich J, Back C H and Jungwirth T 2015 *Rev. Mod. Phys.* 87 1213
[5] Uchida K, Takahashi S, Harii K, Ieda J, Koshibae W, Ando K, Maekawa S and Saitoh E 2008 *Nature* 455 778–81
[6] Zhang W, Han W, Jiang X, Yang S-H and Parkin S S P 2015 Nat. Phys. 11 496–502
[7] Zhou J et al 2019 Sci. Adv. 5 eaau6696
[8] Lee H-Y et al 2019 APL Mater. 7 031110
[9] Elliott R J 1954 Phys. Rev. 96 266–79
[10] Kiss A, Szolnoki L and Simon F 2016 Sci. Rep. 6 22706
[11] Dyakonov M I and Perel V I 1972 Sov. Phys. Solid State 13 3023
[12] Gamou H, Ryu J, Kohda M and Nitta J 2017 Appl. Phys. Express 10 023003
[13] Ryu J, Kohda M and Nitta J 2016 Phys. Rev. Lett. 116 256802
[14] Kumar A, Akansel S, Stopfel H, Fazlali M, Akerman J, Brucas R and Svedlindh P 2017 Phys. Rev. B 95 064406
[15] Akansel S, Kumar A, Behera N, Husain S, Brucas R, Chaudhary S and Svedlindh P 2018 Phys. Rev. B 97 134421
[16] Nan T, Emori S, Boone C T, Wang X, Oxholm T M, Jones J G, Howe B M, Brown G J and Sun N X 2015 Phys. Rev. B 91 214416
[17] Liu L, Pai C-F, Li Y, Tseng H W, Ralph D C and Buhrman R A 2012 Science 336 555
[18] Liu B L, Moriyama T, Ralph D C and Buhrman R A 2011 Phys. Rev. Lett. 106 036601
[19] Lakshmanan M 2011 Phil. Trans. R. Soc. A 369 1280–300
[20] Demasius K U, Phung T, Zhang W, Hughes B P, Yang S-H, Kellock A, Han W, Pushp A and Parkin S S P 2016 Nat. Commun. 7 10644
[21] Ryu J, Avci C O, Karube S, Kohda M, Beach G S D and Nitta J 2019 Appl. Phys. Lett. 114 142402
[22] Tao X et al 2018 Sci. Adv. 4 eaat1670
[23] Wang X, Ortiz Pauyac C and Manchon A 2014 Phys. Rev. B 89 054405
[24] Zimmermann B, Mavropoulos P, Heers S, Long N H, Blügel S and Mokrousov Y 2012 Phys. Rev. Lett. 109 236603
[25] Wang L, Wesselink R J H, Liu Y, Yuan Z, Xia K and Kelly P J 2016 Phys. Rev. Lett. 116 196602
[26] Boone C T, Shaw J M, Nembach H T and Silva T J 2015 J. Appl. Phys. 117 223910
[27] Long N H, Mavropoulos P, Bauer D S G, Zimmermann B, Mokrousov Y and Blügel S 2016 Phys. Rev. B 94 180406(R)