Ultra-low switching current density in all-amorphous W-Hf / CoFeB / TaOx films

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We study current-induced deterministic magnetization switching and domain wall motion via polar Kerr microscopy in all-amorphous W_{66\%}Hf_{34\%}/CoFeB/TaOx with perpendicular magnetic anisotropy and large spin Hall angle. Investigations of magnetization switching as a function of in-plane assist field and current pulse-width yield switching current densities as low as 3 \times 10^9 A/m^2. We accredit this low switching current density to a low depinning current density, which was obtained from measurements of domain wall displacements upon current injection. This correlation is verified by investigations of a Ta/CoFeB/MgO/Ta reference sample, which showed critical current densities of at least one order of magnitude larger, respectively.

I. INTRODUCTION

Switching of ferromagnetic thin films by means of the spin Hall effect (SHE) in heavy metal/ferromagnetic (HM/FM) bilayers with perpendicular magnetic anisotropy (PMA) has been studied intensively over the last years [1–15]. In these systems, the switching current density $j_{sw}$ is typically in the order of $10^{10}$ to $10^{11}$ A/m$^2$. Recent work suggests that the SHE switching is limited by the depinning of domain walls [16]. Therefore, a decrease of the depinning current density $j_{dep}$ should result in a reduced $j_{sw}$. Theoretical descriptions of the depinning field and depinning current density [17, 18], respectively, of skyrmions predict a maximum at $d/(g) \approx 1$, where $d$ denotes the skyrmion diameter and $g$ the average grain size. For larger and smaller grains with respect to the skyrmion diameter, a decrease of the critical driving force is expected. Films containing large grains can be grown epitaxially, which lately yielded low switching current densities in Au/Fe$_3$N bilayers [19]. A decrease of the grain size, on the other hand, can be realized in disordered/amorphous systems. The resulting low pinning of skyrmions was recently demonstrated in W/CoFeB/MgO thin films prepared via sputter deposition with high deposition rates [20]. A demonstration of a low switching current density as a result of reduced pinning due to smaller grains, however, is still pending.

Considering the similarity between Skyrmions and domain walls, the depinning current density of domain walls should exhibit a maximum at $\lambda/(g) \approx 1$ and reduced depinning current densities for larger or smaller grains, respectively. Here, $\lambda$ denotes the domain wall width parameter. Based on this hypothesis, we investigate in the present work if the lack of crystallinity results in smaller switching current densities due to reduced pinning of domain walls. For this purpose, we continue our investigations of a W$_x$Hf$_{1-x}$,1 nm / CoFeB 3 nm / TaOx 2 nm system [21], which exhibits a phase transition from a segregated phase mixture to an amorphous alloy for $x \leq 0.7$. Due to the accompanying jump in resistivity, the SHA shows a pronounced maximum of $\theta_{SH} = -0.2$ at the phase transition. In this system, PMA can be obtained by decreasing the FM layer thickness and post-annealing the sample. First observations of domain nucleation and expansion upon out-of-plane field application showed the formation of large domains with only few pinning sites. We understand this as a hint for weak pinning in the amorphous W-Hf, making this system interesting for investigation of a correlation between the lack of crystallinity and weak pinning. Finally, the large SHA of the amorphous W-Hf should allow for efficient current-induced magnetization switching. Here, we report on current-induced magnetization switching experiments (CIMS) and current-induced domain wall motion (CIDWM) experiments. We study the correlation between switching current density and depinning current density in detail. Additionally, we perform experiments with another material system, which is known to exhibit stronger domain wall pinning. Our CIMS and CIDWM experiments are supported by detailed magnetic characterization of the films and measurements of the spin Hall effective fields in various configurations.

II. METHODS

The thin films were grown in UHV magnetron (co-) sputtering systems at room temperature on thermally oxidized Si wafers and post annealed in a vacuum furnace with a pressure below 5 \times 10^{-7} mbar. W-Hf thin films were prepared via co-sputtering. Thermal stability of the amorphous phase was verified by comparing the sheet resistance, measured via four-probe technique, before and after annealing. A nominal tungsten content of 66 \% was chosen for further investigations. All layers in

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this sample were deposited with an Ar working pressure of \(2 \times 10^{-3}\) mbar. Perpendicular magnetic anisotropy (PMA) was obtained with a thin CoFeB layer and by post-annealing the sample at 180 °C for 20 min. The full stack is Si (001) / SiOx 50 nm / WmgHf3 8 nm / Co80Fe40B20 0.85 nm / TaOx 2.55 nm / SiN 1.5 nm. For comparison, a reference sample consisting of Si (001) / SiOx 50 nm / Ta 8 nm / Co80Fe40B20 1.1 nm / MgO 1.8 nm / Ta 1.5 nm was prepared. The Ar working pressure during the sputter deposition was 1.22 \(\times\) 10^{-3} mbar, and, for the MgO layer, 2.2 \(\times\) 10^{-2} mbar. Here, the usage of MgO allows for PMA with a thicker CoFeB layer. Longitudinal and polar magneto-optical Kerr effect (MOKE) measurements at room temperature were performed to characterize the magnetization of the samples. To ensure comparability of the two systems, the anisotropy of the reference sample was set to a value similar to that of the W-Hf system in dependence on the pulse width resulting hysteresis loops (cf. Fig. 3(a)), as a function of the applied current density. From the two systems, the anisotropy of the reference sample at room temperature were performed to characterize the motion and expansion of the domain walls. Within the walls, the magnetic moments lie in the film plane, therefore we need to investigate the effective fields acting on moments in the film plane. Therefore, we additionally conduct measurements with in-plane geometry. The in-plane measurement scheme was identical to our previous work [21], in which a dual Halbach cylinder array with a rotating magnetic field up to \(B_{\text{ext}} = 1.0\) T (MultiMag, Magnetic Solutions Ltd.) was used for in-plane field rotation. The scheme is facilitated by the weak PMA in our samples. The current-induced effective SOT field amplitudes \(B_{\text{DL}}\) and \(B_{\text{FL}},\) associated with the DL and FL spin-orbit torques, were derived from the second harmonic Hall voltage rms value

\[
V_{2\nu} = \left( - \frac{B_{\text{FL}}}{B_{\text{ext}}} R_{P} \cos 2\varphi - \frac{1}{2} \frac{B_{\text{DL}}}{B_{\text{eff}}} R_{A} + \alpha'I_{0} \right) I_{\text{rms}} \cos \varphi.
\]

The DL effective fields and the anomalous-Nernst contribution \(\alpha'I_{0}\) were separated by their dependence on the external field. Here, \(\varphi\) is the in-plane field-angle with respect to the current direction and \(B_{\text{eff}} = -B_{\text{sHE}} + B_{\text{ext}}\) is the effective magnetic field. We use the convention \(B_{\text{sHE}} > 0\) for spontaneous PMA. From the SOT fields and the current density amplitude \(j_{0}\), the SOT efficiency \(\xi_{\text{DL/FL}} = (2e/h)M_{t}FBM_{t}B_{\text{DL/FL}}/c_{\text{AR}}/j_{0}\) was determined as a weighted mean from individual measurements, taking into account the aspect ratio of the Hall bars with a correction factor \(c_{\text{AR}}\). Assuming that \(\xi_{\text{FL}} = 0\) and that the interfaces are perfectly transparent for the spin current, we have \(\xi_{\text{DL}} = \theta_{\text{SH}}\). Because of the similar resistivities of all conducting layers (about 180 to 200 \(\mu\Omega\) cm), no correction within a parallel resistor model had to be applied.

III. RESULTS AND DISCUSSION

In our previous work [21], we reported on a large SHE in W-Hf/CoFeB/TaOx/SiN films which was observed as a
result of the formation of an amorphous phase for a tungsten content below 70%. Due to the accompanying jump in resistivity, the spin Hall angle shows a pronounced maximum of $\theta_{\text{SH}} = -0.2$ at the phase transition. This stoichiometry, however, showed a decrease in resistivity upon post annealing, indicating crystallization. In order to find a thermally stable amorphous composition with a SHA as large as possible, W-Hf thin films with nominal tungsten content of 62% to 68% were prepared. Thermal stability is verified in all samples and the measured high resistivities confirm equivalence to our previous work. To compensate for slight process instabilities during the deposition, a tungsten content of 66% was chosen for further investigations. Due to the usage of TaOx instead of MgO, which is known to crystallize upon annealing, we assume to maintain the all-amorphous character of our sample stack throughout the sample processing. The final film stack has a resistivity of $\rho_{\text{xx}} = 185 \, \mu \Omega \text{cm}$ and a magnetization of $M_s = 684 \, \text{kA/m}$. The low magnetization indicates the presence of a magnetic dead layer in the CoFeB, probably due to oxidation during the formation of the TaOx layer. From MOKE measurements, presented in Fig. 1(a) and (b), an anisotropy field of $B_{\text{ani}} = 123 \, \text{mT}$ and a very low coercive field of $B_c \approx 80 \, \mu \text{T}$ were determined. The latter allows for nucleation and expansion of domains in the W66Hf34/CoFeB/TaOx/SiN film with low out-of-plane field pulses, as shown in Fig. 2 for field pulses with $B_x = 100 \, \mu \text{T}$ and $\tau = 0.1 \, \text{s}$. The resulting domain only contains with few macroscopically visible pinning sites with distances in the order of $50 \ldots 100 \, \mu \text{m}$.

The Ta/CoFeB/MgO/Ta reference sample was post-annealed to match the anisotropy of the W66Hf34 sample, resulting in $B_{\text{ani}} = 120 \, \text{mT}$ and $B_c = 3 \, \text{mT}$, determined from the MOKE measurements shown in Fig. 1(c) and (d). The final sample stack has a resistivity of $\rho_{\text{xx}} \approx 180 \, \mu \Omega \text{cm}$ and a saturation magnetization of $M_s = 1088 \, \text{kA/m}$.

In Fig. 1(a) we show hysteresis loops obtained by observing CIMS in the $50 \times 6 \, \mu \text{m}^2$ line bars with varying in-plane longitudinal fields in the all-amorphous W-Hf based sample. Each data point was recorded after $N = 20$ pulses with a pulse width $\tau = 1 \times 10^{-3} \, \text{s}$. Analogously, CIMS experiments were conducted with different pulse widths and the resulting $j_{\text{sw}}$ is displayed in Fig. 1(b) as a function of $\tau$ for varying in-plane fields. We observe a decrease of $j_{\text{sw}}$ for increasing pulse width and longitudinal field, with ultra-low switching current densities in the range of $3 \times 10^9 < j_{\text{sw}} < 2.8 \times 10^{10} \, \text{A/m}^2$. To the best of our knowledge, these are the lowest switching current densities reported so far in HM/FM bilayer systems and a similarly low switching current density was found only in an epitaxial system [19]. The same experiment was performed with the Ta reference sample. The resulting $j_{\text{sw}}$, presented in Fig. 1(c), again shows a decrease with increasing field and pulse width and matches the results from other reports [11, 16] found for similar layer stacks. However, with $1.2 \times 10^{11} < j_{\text{sw}} < 3.0 \times 10^{13} \, \text{A/m}^2$ the switching current density of the Ta reference is up to 40 times larger than in the amorphous W66Hf34 sample.

For better understanding of the origin of the low switching current density, we observed the CIMS of the W66Hf34 sample in line bars in the differential Kerr image. As can be seen in figure 4, we find that the magnetization reversal happens via nucleation of a single domain which then quickly expands upon application of current pulses. Thus,
the switching is dominated by the motion of domain walls rather than by the nucleation process \cite{29} and, therefore, the domain wall velocity is the limiting factor for CIMS.

The velocity of domain walls $v_{DW}$ as a function of the driving force, in our CIMS experiment the applied current density, can be generally divided in two regimes, the creep and the flow regime, separated by a region in which the depinning of the domain walls takes place \cite{30,32}. In the creep regime, the motion of domain walls is thermally activated and the velocity can be described by the creep law

$$v_{DW} = v_{dep} \cdot \exp \left( -\frac{T_{dep}}{T} \cdot \left[ \left( \frac{j}{j_{dep}} \right)^{-1/4} - 1 \right] \right) \quad (3)$$

in which $j_{dep}$ denotes the depinning current density and $T_{dep}/T = E_{dep}/k_B T$ corresponds to the thermal stability factor $\Delta$ of the system \cite{33}.

To investigate the relation between $j_{sw}$ and $j_{dep}$, measurements of the domain wall velocity as a function of applied current density were performed and the creep law (eq. 3) was fitted to the data. We define the depinning current density here as the value $j$ where the creep law deviates from the measured data by at least one standard error. For the W$_{65}$Hf$_{34}$ sample the measurements were performed in the $50 \times 6 \mu m^2$ line bars by applying current pulses with $\tau = 5 \times 10^{-7} s$ and varying number $N$. CIDWM is observed for zero assist field, indicating that the domain walls are of a partial Néel-type due to the presence of Dzyaloshinskii-Moriya interaction (DMI) \cite{34}. Without an external magnetic field, up/down and down/up domain walls move in the same direction, indicating that they have the same chirality. The motion of up/down and down/up domain walls was analysed separately. The resulting domain wall velocity for zero assist field, presented in Fig. 5(a), allows for a clear distinction between the creep and the flow regime. From fitting the creep law, we find a low depinning current density of $j_{dep} = (3.9 \pm 0.3) \times 10^{10} A/m^2$ with a thermal stability factor of $\Delta = 27 \pm 4$. We find a linear dependence of $v_{DW}$ in the flow regime, the resulting line fit, however, does not intercept at zero which could be explained by a low Gilbert damping or excessive Joule heating due to the high resistivity of the material. Extrapolating the CIMS results in Fig. 3(b) to $\tau = 5 \times 10^{-7} s$ for $B_x = 10 mT$ we obtain the corresponding switching current density $j_{sw} \approx 2.5 \times 10^{10} A/m^2$, i.e. nearly identical to the depinning current density for the same pulse width. We note that $B_x = 10 mT$ is approximately the threshold value for observation of CIMS, which indicates that only for larger $B_x$ the homochirality of up/down and down/up domain walls is broken. This allows to quantify the strength of the DMI from an individual analysis of $v_{DW}$ for up/down and down/up domain walls as a function of $B_x$ \cite{35,36}. A minimum of $v_{DW}$ is observed for a particular $B_x$, where the sign of this field is opposite for the up/down and down/up
domain walls. Here, the effective longitudinal DMI field is exactly compensated, resulting in Bloch-type domain walls at one side of a domain and, therefore, vanishing $B_{DL}$. From this field, the effective DMI constant

$$D = B_{DMI} M_s \lambda$$  (4)

can be calculated \[35\] \[36\]. Here, $\lambda = \sqrt{A/K_{uv}}$ is the domain wall width parameter in which $A \approx 20 \text{pJ/m}$ denotes the exchange constant \[37\] and $K_{uv} = B_{ani} M_s/2$ is the effective uniaxial anisotropy \[38\]. We observe a minimum $v_{DW}$ for an in-plane assist field of $B_x \approx 10 \text{mT} = B_{DMI}$. With $\lambda \approx 22 \text{nm}$, a DMI constant of $D \approx 0.15 \text{mJ/m}^2$ is obtained. In the Ta reference sample, CIDWM is only observed under application of longitudinal fields, indicating Bloch-type domain walls \[34\], where the magnetic moments in the domain walls are aligned parallel to spin direction $\sigma$ of the spin current. In order for a spin current to cause a DL field $B_{DL} = \sigma \times m$, the magnetic moments have to be tilted by an external field to have a magnetization component perpendicular to $\sigma$. For this experiment, line bars with a geometry of $100 \times 2 \mu\text{m}^2$ were used to prevent nucleation of domain walls at the line edges due to the Oersted field. The resulting domain wall velocity, measured for different longitudinal fields and with $\tau = 5 \times 10^{-7} \text{s}$, is presented in Fig. 3(b). Larger in-plane assist fields force the magnetization within the domain walls to be parallel to the applied current, resulting in a larger $B_{DL}$ and higher DW velocities. The depinning current density is obtained from fitting the creep law to the $10 \text{mT}$ measurement, as this field equals $B_{DMI}$. We find $j_{dep} = (2.9 \pm 0.3) \times 10^{11} \text{A/m}^2$ and a thermal stability factor $\Delta = 105 \pm 7$. The corresponding switching current density is obtained by extrapolation of the data in Fig. 3(c). For $B_x = 10 \text{mT}$ we obtain $j_{sw} \approx 2.5 \times 10^{11} \text{A/m}^2$.

A comparison of the results of both samples emphasizes a direct correlation between $j_{dep}$ and $j_{sw}$, as both current densities are of the same order of magnitude, respectively. Additionally to the lower depinning and switching current densities in W66Hf34-based sample, we observe a significantly lower thermal stability factor, which is in good agreement with earlier investigations of the correlation between the thermal stability and switching current density in CIMS and CIDWM experiments with submicron stripes, where pinning results from edge effects \[35\]. We note that the thermal stability factor is probably underestimated for the Ta-based sample, because we neglect the effect of Joule heating in its calculation. A stationary finite-element analysis of the structures for long pulses indicates that Joule heating is weak in W66Hf44: for the maximum applied current density of $4 \times 10^{10} \text{A/m}^2$, a temperature rise of $\Delta T < 2.4 \text{K}$ is expected at most. In the Ta-based reference sample on the other hand, the calculations yield a temperature rise of $\Delta T < 93 \text{K}$. Thus, the Joule heating plays a more prominent role, which certainly leads to a reduction of the switching and depinning current densities. From the thermal stability factor, the depinning energy $E_{dep} = k_B T \Delta$ is estimated to be $E_{dep}(\text{W66Hf34}) \approx 0.69 \text{eV}$ and $E_{B}(\text{Ta}) \approx 3.5 \text{eV}$. Here, the Joule heating is taken into account as described above.

Measurements of the DL and FL effective fields are performed in both samples to ensure that the significantly lower switching and depinning current densities in W66Hf34 are not the result of a much larger SOT efficiency. For the Ta reference, we find identical SOT efficiencies for both in-plane and out-of-plane measurement schemes. Furthermore, DL and FL values are very similar, incidentally, where the weighted mean is $\xi_{DL} \approx \xi_{FL} \approx -0.05 \pm 0.004$. In the W66Hf34 sample, the out-of-plane measurements scheme yields $\xi_{DL} \approx -0.14 \pm 0.0003$ and $\xi_{FL} \approx -0.02 \pm 0.005$. From the in-plane configuration, we obtain $\xi_{DL} \approx -0.14 \pm 0.02$ and $\xi_{FL} \approx 0.09 \pm 0.03$. Again, the DL torques are very similar in both schemes, whereas the FL efficiencies are clearly different. The efficiency of the DL torque, which gives rise to the current-induced motion of the domain walls is approximately 2.8 times larger in the W66Hf34-based sample. However, this is not sufficient to explain the factor of up to 40 between the switching and depinning current densities of the two sample types. The main difference between the samples is thus clearly the pinning strength, which is very weak in the W66Hf34-based sample.

**IV. CONCLUSION**

In summary, we investigated the current-induced magnetization switching in an all-amorphous W66Hf34/CoFeB/TaOx sample stack, obtaining ultra-low...
switching currents densities as low as $3 \times 10^9\, \text{A/m}^2$. Observation of domain wall motion reveals a depinning current density in the same order of magnitude with a ratio of $j_{\text{dep}}/j_{\text{sw}} \approx 1.6$. A comparison with a Ta/CoFeB/MgO/Ta reference sample yields $j_{\text{dep}}/j_{\text{sw}} \approx 1.2$, emphasising that this ratio is universally close to unity \cite{39}. This result indicates a correlation between the switching and depinning current density, where the switching is limited by the propagation of domain walls, rather than by their nucleation. Observations of the current induced magnetization switching process in the differential Kerr image confirm this conclusion. The ultra-low switching results in reduced pinning. This combination of a low current densities we obtained for W$_{66}$Hf$_{34}$ therefore suggest that the all-amorphous character of the sample results in reduced pinning. This combination of a low pinning and the potentially large spin Hall angle in all-amorphous heavy metals, given by the intrinsic spin Hall effect due to the high resistivity, could be interesting for further investigations in the context of magnetization switching.

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