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Cite as: AIP Advances 10, 015336 (2020); https://doi.org/10.1063/1.5130488
Submitted: 23 October 2019 . Accepted: 08 January 2020 . Published Online: 23 January 2020

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COLLECTIONS

Paper published as part of the special topic on 64th Annual Conference on Magnetism and Magnetic Materials
Note: This paper was presented at the 64th Annual Conference on Magnetism and Magnetic Materials.
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ABSTRACT
We investigated a magnetic nanowire device for a light modulator which is very important for 3D holography display application. It has a Si-N/Ta/Gd-Fe/Si-N nanowire with 0.5×17 μm² dimension and a Co/Pd multilayer hard magnet with its size of 0.5×3 μm² which is attached underneath the nanowire for initial domain nucleation. Both the nanowire and the hard magnet have perpendicular anisotropy. An initial domain was successfully nucleated by pulse current and found spin-orbit torque from Ta and local magnetic field from the nanomagnet was responsible for the domain nucleation. Current induced domain wall (DW) motion was observed and the DW moved along current direction without magnetic field. DW velocity was significantly enhanced with in-plane magnetic field ($H_{\text{inp}}$) of 25 mT and velocity more than 200 m/s was obtained, which was more than 25 times faster than that without magnetic field. We also show that the fast DW velocity is attributed to chiral Neel domain wall induced by external in-plane magnetic field. This fast DW motion is very useful for large scale light modulator which refreshes large number of pixels, such as 10K×10K pixels, over a limited time scale of 8~16.7 ms which corresponds to TV frame rate of 120 or 60 Hz.

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I. INTRODUCTION
Magneto-optical spatial light modulators (MOSLM) driven by spin-transfer switching with current-perpendicular-to-plane (CPP) devices have been developed for 3D holography display application with wide viewing zone. And it has successfully shown display capability with a pixel of 2×2 μm².1,2 A pixel pitch of the SLM as small as 1 μm enabled large viewing angle as wide as 30 degrees, which is very important for 3D spatial images though none of the current SLM achieved so far.3 However, the driving current for the one μm pixel with the CPP device was more than the capability of a cell selection transistor with 1μm pixel pitch, and further reduction of driving current has been pursued. One solution for this issue is current in-plane type MO light modulator, which uses current-induced domain wall motion (CIDWM) to switch the magnetization direction.4 However, driving speed of the device, which is proportional to domain wall (DW) velocity, has to be fast enough to refresh pixel data over large scale such as 10K×10K or larger. Fast CIDWM has been found in a magnetic layer sandwiched by a heavy metal and an insulator. The fast DW motion is attributed to spin orbit torque (SOT) driven DW motion5 with chiral Neel DW structure which is promoted by Dzyaloshinsky-Moriya interaction (DMI). However, there is not much research about how initial domains nucleated on the SOT driven nanowire and its CIDWM for a light modulator. In this article we investigate a Ta/Gd-Fe nanowire with a hard magnet (HM), and show a peculiar domain nucleation and velocity of CIDWM over 200 m/s.

II. SAMPLE PREPARATION AND DEVICE FABRICATION
Fig. 1(a) shows a bird-eye view illustration of a fabricated nanowire device, which comprises a magnetic nanowire of Ta (3 nm)/Gd<sub>22.6</sub>-Fe<sub>77.4</sub>(10-15 nm)/Si-N(5 nm) with 500 nm width, a
HM of [Co(0.3 nm)/Pd(0.6 nm)]_{25} with 0.5 × 3 μm², and electrodes for current injection. The HM is embedded in the SiN trench, and the magnetic nanowire is fabricated on top of the HM and SiN. Fig. 1(b) shows a micrograph of the fabricated device. We fabricated the devices with three types of nanowires, (sample A’): Si-N/Gd-Fe-22.6 at% (15 nm)/SiN, (sample A): Si-N/Gd-Fe-22.6 at% (15 nm)/SiN, (sample B): Si-N/Ta (3 nm)/Gd-Fe-22.6 at% (15 nm)/SiN, and (sample C): Si-N/Ta (3 nm)/Gd-Fe-22.6 at% (10 nm)/SiN. Films of the NMs and the nanowires were deposited by a magnetron sputtering system. The composition of the Gd-Fe was controlled by co-deposition of Gd and Fe targets changing each target power and was confirmed by X-ray Fluorescence method. Those films were patterned by an electron beam lithography and ion beam milling followed by Si-N deposition using ion beam sputtering system and liftoff process.

III. INITIAL DOMAIN NUCLEATION

Definition of the coordinate is shown in Fig. 1(a), positive current along +x-direction, and magnetic field is applied z- (H_z) and x-direction (H_x). Figure 2(a) shows Kerr hysteresis loops of the sample A (without Ta layer) showed the switching field in negative field was smaller than that in the positive field, while the sample B and C (with Ta layer) showed same switching field of the negative/positive field. This asymmetric switching field behavior in the samples A’ and A is explained by the different switching mode. The small switching field at negative field is due to domain propagation from an initial domain. We will discuss more detail later.

We think the switching in samples B and C is coherent rotation because of their large switching field, and do not know the reason of the symmetrical switching though they have almost same HM structure except for the Ta layer under the Gd-Fe. As you can see the Kerr rotation angle of the sample A is quite different from that of the sample B even though they have similar film structure. We think this is insubstantial, because the Kerr rotation angle is calibrated by assuming the uniform reflectance of the sample. They cannot be evaluated quantitatively, especially when the patterned sample (0.5 μm) is smaller than the beam spot (3 μm). And a slight off-alignment of the sample from the beam center can also change the Kerr rotation magnitude.

Figure 2(b) shows MO micrographs (taken by differential mode) at zero magnetic field (initial state) after applying +800 mT. Initial magnetic domains were observed for the sample A’ and no initial domain was observed for other samples. We think that the domain was nucleated by the local magnetic field of the HM and is responsible for the small switching field with negative field with sample A’. We speculated that slightly larger negative field is required for domain nucleation for the sample A. This explanation is consistent with the larger negative switching field of the sample A compare with that of the sample A’. The large switching field at positive magnetic field for the samples A’ and A is due to coherent rotation since no initial domain was nucleated.

Figure 2(c) shows MO micrographs after current pulse injection for initial domain nucleation, which were observed as dark spots beside the HM. Initial domains were observed on both side of the HM in the sample A’ and A, while it was observed on the left (right) side for the sample B and C when the positive (negative) current pulse was applied.

Figure 3 shows a cross-sectional illustration of the device with directions of magnetic fringe field from the HM and current to explain how the initial domain nucleated. H_x and H_y represent x- or z-component of the fringe field. The initial domain nucleation near the HM in the sample A can be explained by local magnetic field from the HM (H_y) and thermal coercivity degradation of the Gd-Fe induced by current injection. On the other hand, the peculiar nucleation in the sample B and C is qualitatively attributed to the SOT with H_x (which is known for the SOT switching) in addition to the H_y and the thermal degradation. As shown in Refs. 11 and 12, SOT with in-plane x-direction magnetic field assist the magnetization switching from down to up, while that with y-direction magnetic field prevent the magnetization switching. It is consistent to consider SOT induced magnetization switching attributed our initial nucleation results in the sample B and C.

IV. CURRENT INDUCED DOMAIN WALL MOTION

We evaluated the CIDWM using multiple current pulses with pulse duration of 15 ns after we move the DW to the center of the
FIG. 2. (a) Micro Kerr hysteresis loops for the samples A(A') to C. (b) Initial state of MO micrographs for the sample A(A') to C, taken after +800 mT for saturation. (c) MO micrographs after current injection for initial domain nucleation. Current amplitude was 28 MA/cm$^2$ for the sample A(A'), 37 MA/cm$^2$ for the sample B and 23 MA/cm$^2$ for the sample C. Pulse width was 50 ns.

The position of the DW was realized by multiple current pulse injections after an initial domain nucleation. Figure 4 shows MO micrographs of sample A and C before and after current injection of $-2.2\text{mA}$ (pulse width of 15ns 30 times) without magnetic field. The DW of the sample A moved along with electron flow direction, while that of the sample C (also B) moved along current flow direction (the sample B was not shown).

Figure 5 shows current density versus DW velocity of the sample A with/without in-plane magnetic field ($H_{\text{inp}}$) for expecting faster CIDWM. The velocity was determined by DW displacement divided by total pulse duration. The DW velocity of the sample A increased with an increase with current amplitude, however it became slower when the $H_{\text{inp}}$ was applied. On the other hand, the DW velocity of the sample C significantly increased with the $H_{\text{inp}}$ as shown in Fig. 6. The velocity more than 200 m/s was realized with $H_{\text{inp}} = 25\text{mT}$, while the velocity was less than 10 m/s without $H_{\text{inp}}$. And when the direction of the $H_{\text{inp}}$ reversed, the direction of the up/down DW motion reversed (Fig. 6(a)). The DW motion direction of the up/down DW is opposite to that of the down/up DW with same $H_{\text{inp}}$ (Fig. 6(b)).

Most of our results are explained by the efficient DW motion by SOT, which is originating from the chiral Neel DW with spin accumulation in heavy metal. The fast DW velocity with $+H_{\text{inp}}$ is explained by the chiral Neel DW promoted by $H_{\text{inp}}$, while the small DW velocity without $H_{\text{inp}}$ is attributed to poor chiral Neel DW or Bloch DW due to insufficient DMI. Figure 3(b) shows a schematic illustration how the up/down Neel DW was promoted by $H_{\text{inp}}$ and driven by the SOT. Effective magnetic field induced by the SOT ($H_{\text{SOT}}$) was described by eq. 1 with magnetization direction $\mathbf{m}$ and spin accumulation $\mathbf{s}$. The effective field points toward $-z$ direction at the middle of the DW when the $H_{\text{inp}}$ and current is $+x$.  

FIG. 3. (a) A cross-sectional illustration of the device with direction of local magnetic field from the hard magnet. (b) A cross-sectional illustration of the device explaining how the chiral Neel domain affected by spin orbit torque.

FIG. 4. MO micrographs before and after current injection for (a) the sample A, and (b) the sample C.

FIG. 5. Domain wall velocity versus injected current for sample A with/without in-plane magnetic field.

The field turns the magnetization right handed chiral direction, which drives the DW toward -x direction (opposite to the current direction). The $H_{SOT}$ points to +z direction when the $H_{inp}$ is -x direction and drives the DW to +x direction. However, when the DW is the Bloch wall, no effective field is promoted since $m$ is parallel to the $s$. These theoretical investigations agree with our experiments very well.

\[
\vec{H}_{SOT} \propto (\vec{m} \times \vec{s})
\]  

One should note that DW velocity as fast as a few tens to over one hundred of m/s was obtained at the 1 mA with which showed very slow DW motion without the in-plane magnetic field. These low current driving and fast DW motion are very important for MOSLM device integration, since the small driving current is required for being driven by small backplane transistors and DW velocity has to be fast enough to be driven by short pulse for video frame rate.

Figure 7 shows DW velocity versus $H_{inp}$ for the sample C. As shown in Fig. 7(a) the up/down DW velocity with $-2.2$ (+2.2) mA linearly decreased (increased) with an increase in $H_{inp}$, while down/up DW velocity with $-2.2$ (+2.2) mA linearly increased (decreased) with an increase in $H_{inp}$. The results of Fig. 7 show that the DW with left handed chirality moves along electron flow direction while the DW with right handed chirality moves along current flow direction. These results are very consistent with the results of Ta/CoFeB/MgO.\(^{14}\) We define in-plane magnetic field, $H_L$, which gives zero DW velocity shifted negative (positive) field direction for the up/down (down/up) DW. The effective DMI field, $H_{DMI}$, is defined as $H_{DMI} = -H_l$ which the cancel $H_L$.\(^{8,9,13,14}\) We calculated the $H_{DMI}$ with DW velocities less than 12mT for choosing better linear region using least-square method and obtained $H_{DMI} = 1.1$ and $-1.4$ (mT) for down/up- and up/down-DW, respectively. Polarity of the $H_{DMI}$ is consistent with the previous results,\(^{14}\) however our results are much smaller. We think the weak DMI field is responsible for the slow DW velocity without magnetic field because of their insufficient Neel DW formation. We attributed the small
FIG. 6. Domain wall velocity versus injected current for sample C with/without in-plane magnetic field. (a) Up/down DW (b) down/up DW.

FIG. 7. Domain wall velocity versus in-plane magnetic field for sample C with (a) up/down DW (b) down/up DW.

$H_{DMI}$ to the thick (10 nm) Gd-Fe compared to the reference results since the DMI for unit volume can be reduced for thick magnetic film.

V. CONCLUSION

We have fabricated a Si-N/Ta/Gd-Fe/Si-N nanowire device having a patterned hard magnet underneath and investigated initial domain nucleation and current induced domain wall (DW) motion. The DW nucleation was very different from that with nanowire without Ta, and the nucleation was attributed to spin orbit torque with local magnetic field from the hard magnet. Current induced domain wall motion was significantly enhanced by applying in-plane magnetic field, and we obtained DW velocity more than 200 m/s which is more than 25 time faster compared with that without in-plane magnetic field. For our device having relatively thick Gd-Fe 10 nm, applying in-plane magnetic field is very important for fast DW motion beside just adding a heavy metal.

REFERENCES
