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Spin-orbit torque induced magnetization switching in heavy metal/ferromagnet multilayers with bilayer of heavy metals

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ABSTRACT

Symmetry breaking provides new insight into the physics of spin-orbit torque (SOT) and the switching without a magnetic field could lead to significant impact. In this work, we demonstrate the robust zero-field SOT switching of a perpendicular ferromagnet (FM) layer where the symmetry is broken by a bilayer of heavy metals (HMs) with the strong spin-orbit coupling (SOC). We observed the change of coercivity value by 31% after inserting Co_2FeAl in the multilayer structure. These two HM layers (Ta and Pt) are used to strengthen the SOC by linear combination. With different angles between the magnetization and the current (i.e. parallel and anti-parallel), the structures show different switching behaviors such as clockwise or counterclockwise.

1. Introduction

The ability to switch the magnetization of a single magnetic layer at room temperature using an in-plane current through spin–orbit torque (SOT) opens the way to a new generation of spintronic devices [1–3]. SOT is usually observed in magnetic bilayers made of a heavy-nonmagnetic metal (HM) with strong spin-orbit coupling (SOC) and a ferromagnet (FM) [4]. The charge current passing through the HM layer produces effective fields by two known mechanisms: spin Hall Effect (SHE) and Rashba effect [5–7]. The charge current passing through the HM layer forces a non-equilibrium spin accumulation at the HM/FM interface in the transverse direction via SHE [8,9]. The SOT has drawn increasing interest because it can be utilized in spintronic devices for current-induced magnetization switching [2,3,10], and for high-speed domain wall motion [11,12]. Magnetization switching induced by an in-plane current has been investigated mainly in HM/FM/Oxide(O_x) heterostructures, such as Pt/Co/AlO_x [2,6], Hf/CoFeB(CFB)/MgO and Hf/CFB/TaO_x [13], Pt/$Co_2FeAl_{0.5}Si_{0.5}$ (CFAS)/MgO [14], and also has been shown to work in all-metal structures such as $Mn_{1.5}Ga$/Ta and $Mn_{1.5}Ga$/Pt [15], asymmetric Pt/CoNiCo/Pt [16] and Pt/Co/Pt [17,18] multilayers. Current applied parallel to the interface plane in the trilayers with structural inversion asymmetry generates two qualitatively different types of spin torques, namely a transverse (field-like) torque $\sim\mathbf{m}\times\mathbf{y}$ and a longitudinal (antidamping- like) torque $\sim\mathbf{m}\times(\mathbf{y}\times\mathbf{m})$ [4,5,7,19,20], where $\mathbf{y}$ denotes the in-plane axis perpendicular to the current direction $\mathbf{x}$, and $\mathbf{m}$ is the magnetization unit vector. These two effects operate in distinct disorder regimes, in which the SHE necessitates strong disorder and the nature of the torque may dramatically change from one regime to the other [21–23]. The sign of magnetization switching may reverse between the samples with Pt and Ta under layers due to their opposite sign of spin Hall angle [10]. On the other hand, a small symmetry-breaking bias field is usually needed for the SOT induced magnetization switching [16,24,25]. However, the most critical drawback of SOT is that it requires external in-plane magnetic field for a deterministic switching, which is undesirable for technological applications. Here we investigate the SOT-induced switching in an HM/HM/FM/O_x and HM/HM/FM/FM/O_x multilayer system in the absence of magnetic field. To do so, we are trying several approaches such as changing under HM layer, FM layer or the geometry of the device and using different treatments like etching treatments. The experimental results show that the bilayer HM (Ta and Pt) in multilayer structures has a critical role in determining the torques generated by an in-plane current. In addition, we obtain strong perpendicular magnetic anisotropy (PMA) and strong SOC due to the linear combination of two materials. The strong SOC plays an important role for SOT induced magnetization switching.

2. Experimental sections

Thin films were prepared by magnetron sputtering system without breaking the vacuum of sputtering chamber, Ta (1)/AlO_x (3)/Co (1)/Pt

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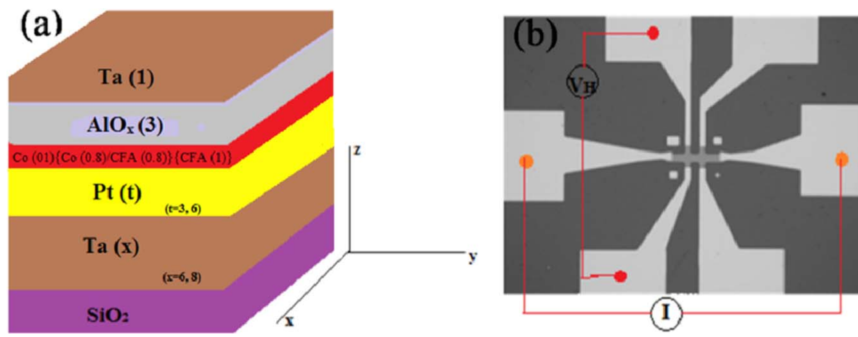

Fig. 1. (a) Film structure of the multilayer. (b) The SEM image of the Hall bar geometry and measurement setup.

(t)/Ta (x), Ta (1)/AlO_x (3)/Co (0.8)/CFA (0.8)/Pt (t)/Ta (x) and Ta (1)/AlO_x (3)/CFA (1)/Pt (3)/Ta (8) multilayers (thickness in nanometer and from top to bottom, here x=6, 8 and t=3, 6) were deposited on Si/SiO_2 substrate as shown in Fig. 1a. These seven devices are referred to as Co device for x=8 and t=3, Co-I device for x=8 and t=6, Co-II device for x=6 and t=3, Co/CFA device for x=8 and t=3, Co/CFA-I device for x=8 and t=6, Co/CFA-II device for x=6 and t=3, and CFA device in the following. The thin films were patterned in three steps with electron-beam lithography (EBL) and two steps with Ar ion milling. Firstly the stack Ta (5 nm)/Pt (55 nm) was patterned into a cross by EBL. Then, the film was patterned into a Hall bar and followed by an Ar ion milling technique. Ar ion milling was performed at the base pressure of $2*10^{-4}$ Pa, and then a stack of Co, Co/CFA, and CFA were deposited. Finally, the electrode contact pads were patterned using EBL and followed by Ar ion milling and deposited of Ta (5 nm)/Pt (70 nm). The geometry of the Hall bar was 17 μm in width and 108 μm in length as shown in Fig. 1b. Ar ion milling treatment ensures the chemical stability of multilayers and preserves the strong PMA typical of Co–Pt/Ta interfaces.

All deposition processes and transport measurements were performed at room temperature. After deposition, we investigate the electrical and magnetic properties of thin films. The electrical and magnetic properties of the device measurements setup and the Hall bar structure are shown in Fig. 1b. The Hall resistance vs magnetic field were measured by Gaussmeter (GM) Hall Effect measurement system with the Keithley 2182 A nanovoltmeter and 2400-C source meter by applying a constant DC current source while monitoring the Hall voltage. The magnetic moment vs magnetic field were measured by the alternating gradient magnetometer (AGM) system by applying 1 T out-of-plane magnetic field. The R-H and M-H hysteresis loops were recorded at room temperature with external magnetic field sweeping from positive to the negative direction along normal to the film plane. The contact pad sheet resistance values were typically 100-600 Ω for Co, Co-I, Co/CFA and Co/CFA-I devices, which are the Ohmic contact. However, the sheet resistances of Co-II and Co/CFA-II devices were recorded as in mega Ohm (MΩ). The current-induced magnetization switching was measured by using a combination of Keithley 6221 and 2182 A devices. A pulsed DC. current with a duration of 0.2 ms was applied to the microwires and the Hall voltage was measured simultaneously. Here, to calculate the charge current density, we simply assumed the current is homogeneously distributed within the cross-section of 12 nm × 17 μm and 12.6 nm × 17 μm for Co and Co/CFA device, respectively. During electrical measurements, the thin films were kept away from the regional field point to avoid residual magnetic field. The mechanism was important to avoid small residual magnetic field during the measurements.

3. Results and discussion

The M-H hysteresis loop of Co, Co/CFA, and CFA device were measured by applying of an out-of-plane field as shown in Fig. 2a, b and c respectively. From the M-H measurements, we observed different magnetic moment values recorded for device Co, device Co/CFA, and device CFA, respectively as shown in Fig. 2d. We observed a small magnetic moment for CFA device with the saturation field about ~0.2 T. For the Co and Co/CFA device we observed large magnetic moment as compared with CFA device because Co/CFA device are composed of two FM layers. These two devices recorded the same saturation field value ~0.0274 T. Hence, these FM bilayers are enhancing strong interaction in the system. Both Co and Co/CFA devices displayed good PMA in all the measurements with small coercivity and anisotropy fields without annealing. On the other hand, the Co and Co/CFA devices have square loops along the perpendicular axis and we ascribed that to the enhanced crystalline orientation.

The Hall resistance was measured as a function of the perpendicular magnetic field under different input currents as shown in Fig. 3a, b, c, d and e. When we applied current ranging from −1 mA to 1 mA the coercivity field was recorded at ~64 Oe for Co device as shown in Fig. 3f, g. From Fig. 3d and e, the coercivity of Co-I and Co/CFA-I devices were recorded at about ~48 Oe and ~133 Oe, respectively. However, we could not observe PMA for both Co-II and Co/CFA-II devices. Similarly, we observed the coercivity field of Co/CFA device about ~90 Oe for small source current ranging from −1 mA to 1 mA as shown in Fig. 3f and g. When we increase the input current ranging from −7 mA to 7 mA for both Co and Co/CFA devices, we observed a small reduction in the coercivity value of the composite structures by 3.1 Oe/mA and 4.5 Oe/mA respectively as shown in Fig. 3f and g. When the source current is −3 mA for both Co and Co/CFA devices, we observed the coercivity field shift by 23.6 Oe and 15 Oe respectively, as shown in Fig. 3f and h. These sudden changes of coercivity were from the inter-layer interaction of two FM materials in the system. From Fig. 3h, we observed the change in coercivity ($\mathbf{\Delta H_c}$=$\mathbf{H_c}(I_+)$ - $\mathbf{H_c}(I_-)$) with the applied current. Current dependence coercivity for Co and CFA devices showed similar behavior. With increasing the source current value, the maximum changes in coercivity were recorded as $\mathbf{\Delta H_c}$~23.6 Oe and $\mathbf{\Delta H_c}$~15 Oe for Co/CFA and Co device, respectively. Fig. 3c shows that the CFA device could not get good PMA with different source current, while further increasing of external field value we could not observe saturation point and they became open hysteresis loops at negative field point. Hence, the result indicates that CFA devices are soft magnetic materials as they become saturated at small values of an external magnetic field. From Fig. 3b, f and g, we observed the change of coercivity value of Co/CFA device by 31% after inserting Co_2FeAl, and we also observed these changes from R-H curves in the same manner. When the source current was increased ranging from 3 to 7 mA, the coercivity field decreases linearly for both Co and Co/CFA devices as expected, while much larger current will break the film.

According to the reference [26], the current dependence of coercivity change was due to Joule heating effect, perpendicular anisotropy of the thin film and influences the magnetic domain structure. The strong dependence of the coercivity on the channel current indicates a significant role of current assisted magnetization switching of the

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