

Electrical Control of Metallic Heavy-Metal–Ferromagnet Interfacial States

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Voltage-control effects provide an energy-efficient means of tailoring material properties, especially in highly integrated nanoscale devices. However, only insulating and semiconducting systems can be controlled so far. In metallic systems, there is no electric field due to electron screening effects and thus no such control effect exists. Here, we demonstrate that metallic systems can also be controlled electrically through ionic rather than electronic effects. In a Pt/Co structure, the control of the metallic Pt/Co interface can lead to unprecedented control effects on the magnetic properties of the entire structure. Consequently, the magnetization and perpendicular magnetic anisotropy of the Co layer can be independently manipulated to any desired state, the efficient spin torques can be enhanced about 3.5 times, and the switching current can be reduced about one order of magnitude. This ability to control a metallic system may be extended to control other physical phenomena.

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I. INTRODUCTION

The heavy-metal–ferromagnet–oxide (HM-FM-oxide) structure is one of the most important building blocks of spintronics, where both the HM-FM and FM-oxide interfaces result in various emergent magnetic phenomena, such as tunnel magnetoresistance [1–4], perpendicular magnetic anisotropy (PMA) [1–8], Rashba spin-orbit coupling (RSOC) [9–12], and the Dzyaloshinskii-Moriya interaction (DMI) [13,14]. In particular, PMA is of great importance for future generations of high-density memory and logic applications. It has been demonstrated that PMA can be controlled by an electric field (EF) applied to the oxide layer [1,3,5,15,16], thereby opening a potential way to achieve ultralow energy magnetization switching with very little Joule heating. Although both theoretical [6,17] and experimental [2,7,8] studies indicate that both interfaces contribute to PMA simultaneously, all EF-control studies only focus on control of the FM-oxide interface where EF can penetrate. The HM-FM interface at which no EF exists has not been explored.

There are two types of EF-control effects. One is the voltage-controlled magnetic anisotropy (VCMA) effect with an electronic origin, where PMA can be modified by the EF-induced redistribution of electron densities among different d orbitals in FMs [5,18–20], or through an EF-modified RSOC and DMI [9,10]. In VCMA, the magnetic anisotropy field of the FM layer can be effectively modified by an external EF [1,5,18,19,21,22], which has been successfully employed to achieve efficient switching

in magnetic tunnel junctions [1,3], modulation of domain-wall propagation [22], and control of the Curie temperature [21] and Skyrmion bubbles [23]. The second type is of an ionic origin, where the ionic motion of O^{2-} driven by an EF is exploited to control the interfacial oxidation states of a FM [15,16,24–26]. Through this solid-state electrochemical reaction at the FM-oxide interface, the saturation magnetization (M_s) of the FM can also be controlled. The ionic effect provides a much larger control over PMA, on the order of 10 pJ/Vm, as opposed to a few hundreds of fJ/Vm with VCMA [15,16]. Moreover, unlike electrons, ions can diffuse into a metallic system even without EFs, and thus the ionic effect may be employed to control the metallic HM-FM interface. It is known that, in addition to PMA, the HM-FM interface also determines spin-orbit-torque- (SOT) driven magnetic dynamics [27,28] and DMI effects [13,14] in these structures. Therefore, it would be highly advantageous if the HM-FM interface could also be controlled electrically, which will pave the way for an alternative to control PMA and simultaneously modulate the interfacial phenomena to improve the efficiency of SOT switching. However, the voltage-controlled HM-FM interface has yet to be demonstrated.

Here, we demonstrate voltage control of the HM-FM interface in a Pt/Co/GdO_x structure. By combining independent control of the Co/GdO_x interface, we reach an unprecedentedly effective manipulation of PMA in HM-FM-oxide structures, greatly expanding the controllable magnetic states to those that cannot be achieved by controlling only the FM-oxide interface in previous studies. Moreover, both the PMA field (H_k) and M_s can be controlled independently and the magnetism can be configured to any possible state, in sharp contrast to previous

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studies where M_s and H_k are always controlled together when only the FM-oxide interface is modulated [15,16]. The efficiency of SOTs is also controlled by voltage, and the critical SOT-switching current (I_c) is reduced by about one order of magnitude. These results highlight the unexplored voltage control of metallic HM-FM interfaces in the electrical control of magnetism, which may be extended to control other interfacial effects [29–31].

II. METHODS AND MATERIALS

A. Sample fabrication

Three batches of Pt/Co/GdO_x samples with structures of Pt (10 nm)/Co (4 nm)/GdO_x (50 nm), Pt (3 nm)/Co (0.9 nm)/GdO_x (80 nm), and Pt (3 nm)/Co (0.6 nm)/GdO_x (80 nm) are deposited on silicon substrates with a 300-nm thermally oxidized SiO₂ layer by magnetron sputtering at a base pressure of 2×10^{-8} torr. The Pt (10 nm)/Co (4 nm)/GdO_x (50 nm) samples with thick Pt and Co layers are used for observing EF-driven oxygen migration under cross-sectional scanning transmission electron microscopy (STEM). The Pt (3 nm)/Co (0.9 nm)/GdO_x (80 nm) and Pt (3 nm)/Co (0.6 nm)/GdO_x (80 nm) samples that show PMA in the as-deposited state with much thinner Co layers are used for investigating EF control of PMA and SOT switching, respectively. All metallic layers are dc sputtered and GdO_x is reactively sputtered. The deposited continuous multilayers are then patterned into Hall bar structures. The residual areas are protected by 100-nm SiO₂. The top gate electrodes are made of Ta (5 nm)/Ru (100 nm).

B. Electrical measurements

The magnetization and H_k are detected with the anomalous Hall effect. All the Hall resistances (R_H) are detected

by applying a 50- μ A dc current. For the SOT switching, a 1-ms current pulse with the desired amplitude is applied to switch the magnetization, and then R_H is recorded. All electrical measurements are performed after the samples cool down to room temperature if heated.

C. STEM measurements

Cross-section STEM samples are prepared using a Zeiss Auriga focused-ion-beam system. STEM imaging is performed on an FEI Titan with CEOS probe aberration corrector operated at 200 kV with a probe convergence angle of 24.5 mrad, spatial resolution of 0.08 nm, and probe current of ~ 20 pA. The STEM samples are loaded into the microscope immediately after the focused-ion-beam sample preparation to avoid possible oxidation of the Co in air.

III. RESULTS AND DISCUSSION

A. STEM observation of voltage-driven oxygen migration in metallic systems

To directly observe the EF-modulated metallic HM-FM interface through ionic effects, we perform STEM and electron energy loss spectroscopy (EELS) measurements in a 4-nm Co layer with the structure of Si/SiO₂ (300 nm)/Pt (10 nm)/Co (4 nm)/GdO_x (50 nm). As shown in Fig. 1(a), the effective voltage-control area is patterned into a $200 \times 200 \mu\text{m}^2$ square. Figure 1(b) shows R_H as a function of the applied vertical magnetic field (H_z) at different voltage-control states. The as-deposited state shows an in-plane anisotropy state [sample *c* in Fig. 1(b)], which could be modified to a PMA state (samples *d* and *f*) or a fully oxidized state (sample *e*) by applying a gate voltage (V_G) [15]. For samples *d* and *e*, only $V_G = -5$ V is applied. For sample *f*, a $V_G = -5$ V is applied first until a fully

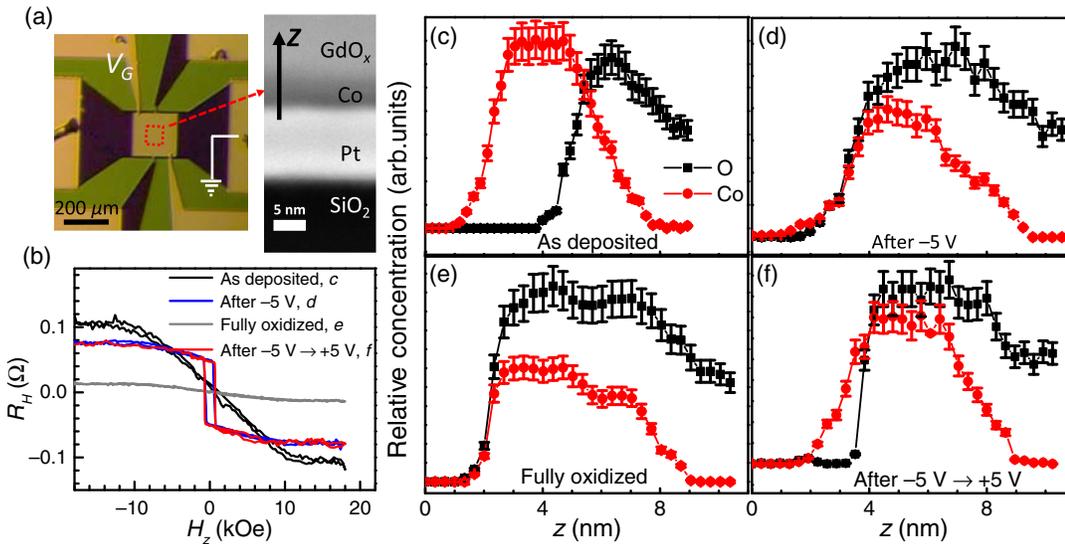


FIG. 1. STEM EELS results at different EF-controlled states. (a) Top view and typical STEM image of a Pt/Co/GdO_x sample. (b) The vertical field-dependent R_H for samples *c*, as-deposited state, *d*, after $V_G = -5$ V application, *e*, fully oxidized state, and *f*, after $V_G = -5$ V and then $V_G = +5$ V applications. (d)–(f) The corresponding Co and oxygen distribution from EELS line scan along the z direction at the Pt/Co interface.

oxidized state is achieved, and then a $V_G = +5$ V is applied to rebuild PMA as shown in previous works [15]. Figures 1(c)–1(f) show the depth distribution of oxygen and Co near the Pt/Co interface. For the as-deposited state [Fig. 1(c)], oxygen ions (O^{2-}) only distribute near the Co/ GdO_x interface and the Pt/Co interface is clean (no O^{2-} detected). After $V_G = -5$ V application, the Co and oxygen curves have overlapped each other at the Pt/Co interface [Fig. 1(d)], but the electrical measurement still shows a clear R_H signal from Co metal at this state [Fig. 1(b)], indicating that the Co is only partially oxidized and O^{2-} can penetrate the unoxidized Co metal and reach the Pt/Co interface. Therefore, the combination of both STEM and electrical measurement results can directly confirm that the EF-driven O^{2-} can diffuse into the Co metal layer. For a fully oxidized state, the Co and oxygen curves [Fig. 1(e)] are similar to Fig. 1(d), but there is no R_H signal detected [Fig. 1(b)], indicating a fully oxidized Co layer.

Figure 1(f) shows the Co and oxygen distributions for sample *f*. As mentioned above, this sample is fully oxidized first by applying $V_G = -5$ V and then is reduced to a PMA state by applying $V_G = +5$ V. Although both samples *d* and *f* show a PMA state with the same R_H , the oxygen signal sharply drops to almost zero before reaching the Pt/Co interface in sample *f*, in contrast to sample *d* in which the oxygen signal drops together with the Co signal at the Pt/Co interface. By comparing the oxygen distributions in these two samples, we can conclude that sample *f* has a much cleaner Pt/Co interface than sample *d*. Since the Pt/Co interface is only manipulated by applied V_G in both samples, these results indicate that the metallic Pt/Co interface can be controlled through EF-driven O^{2-} motion. A slight oxygen signal at

$z \approx 2$ nm in Fig. 1(f) may be due to a small amount of trapped oxygen [25] that cannot be removed under $-V_G$.

B. Schematic of voltage-controlled magnetic states at HM-FM interfaces

Voltage control of metallic HM-FM interfaces must be also accompanied by the control of FM-oxide interfaces in HM-FM-oxide structures, and, moreover, the alteration of FM-oxide interfaces may dominate the overall control effects. Therefore, determining the contribution from HM-FM interfaces during the control process becomes a critical step in investigating the voltage control of HM-FM interfaces. A schematic of the voltage control of metallic HM-FM interfaces is presented in Fig. 2. Here, we assume that the PMA originates from both the Co/ GdO_x interface [4,32,33] and Pt/Co interface [2,6–8,17] in Pt/Co/ GdO_x structures. As observed in Fig. 1, O^{2-} can penetrate the metallic Co layer and reach the Pt/Co interface before the Co is fully oxidized. Figures 2(a)–2(e) schematically show the voltage-gated O^{2-} migration by considering O^{2-} diffusion. The expected perpendicular component of magnetization (M_z) under both in-plane and vertical fields is shown in Figs. 2(f)–2(j), respectively. We classify the O^{2-} in this system into two different types, depending on the strength of the bonding energy between O^{2-} and metal atoms. Type I is the O^{2-} with a weak bond to metal atoms and can be moved along grain boundaries through self-diffusion or be driven by an electric field [25,33–36]. Type II is the O^{2-} strongly bonded to metal atoms, and the significant migration of this type of O^{2-} may require a strong electric field and an elevated temperature [15,37].

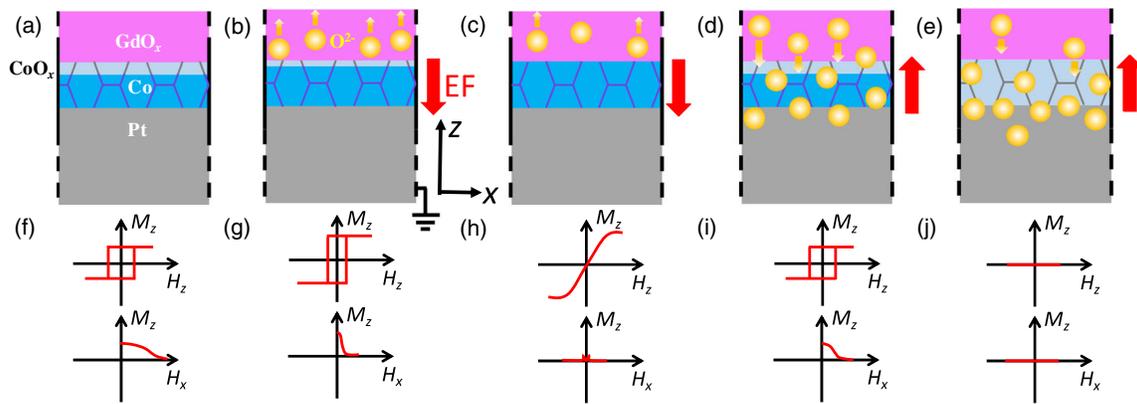


FIG. 2. Schematic of the voltage control of magnetism through metallic Pt/Co interfaces. (a) The as-deposited state showing PMA without V_G application. The Co layer has been partially oxidized and the Pt/Co interface is clean. (b) Under $+V_G$ application, the O^{2-} at the Co/ CoO_x or Co/ GdO_x interface is driven into the GdO_x layer. The CoO_x is partially reduced to Co and the Pt/Co interface keeps clean. (c) The CoO_x is finally reduced to Co and the Co layer becomes in-plane anisotropic under the $+V_G$ application. (d) After $-V_G$ application, the O^{2-} is driven back to the Co/ GdO_x interface and further contaminates the Pt/Co interface through diffusion. PMA is rebuilt and the Co layer is oxidized. Because of the contamination of the Pt/Co interface, H_k becomes weaker compared to (b) even with the same other conditions. (e) The Co layer is fully oxidized by $-V_G$ application. The blue line texture in the Co/ CoO_x layer symbolically illustrates grain boundary. (f)–(j) The schematic of expected vertical and in-plane field-dependent M_z at each state.

As shown in Fig. 2(a), in the initial state, the Co layer is partially oxidized but the Pt/Co interface is clean. This state usually shows a strong H_k because of the orbital hybridization between Co and type-II O^{2-} at the top surface of the Co layer as well as the clean Pt/Co interface. When applying a $+V_G$, the O^{2-} is driven out of the Co/CoO_x or Co/GdO_x interface as shown in Fig. 2(b), resulting in a decreasing H_k and an increasing M_s through the reduction of Co from CoO_x [15]. With increasing $+V_G$ duration, all the O^{2-} can be depleted at the Co/GdO_x interface and the entire Co layer becomes fully reduced to an in-plane anisotropic pure Co metal [15,37] [Fig. 2(c)]. Then, when $-V_G$ is applied, the O^{2-} is driven back to the Co/GdO_x interface [Fig. 2(d)]. Type-II O^{2-} is formed at the Co/GdO_x interface and PMA is recovered. However, by considering the diffusion of type-I O^{2-} through grain boundaries, the situations of Figs. 2(d) and 2(b) will be different. It is expected that type-I O^{2-} can penetrate the Co layer, and thus the Pt/Co interface in Fig. 2(d) will be contaminated by type-I O^{2-} . By considering oxygen contamination at the Pt/Co interface and the contribution of the Pt/Co interface to the PMA, the PMA of Fig. 2(d) will be much weaker than that in Fig. 2(b) with a clean Pt/Co interface, even though the other conditions are exactly the same. Figures 1(d) and 1(f) have directly confirmed the O^{2-} distribution in the two dramatically different situations, and together with magnetic measurements below, these results will provide clear evidence of electrical control of metallic Pt/Co interfaces that has not been explored before. The Co layer can be fully oxidized for a longer $-V_G$ application time [Fig. 2(e)]. If $+V_G$ is then applied again, type-I O^{2-} at the Pt/CoO_x interface will be moved first, and then the CoO_x layer will be gradually reduced. Since type-I O^{2-} has been removed near the Pt/CoO_x interface before the reduction of Co, the sample returns to the state of Fig. 2(b). Because M_s mainly depends on the oxidization and reduction of the Co layer at the Co/GdO_x interface while H_k is determined by both the Pt/Co and Co/GdO_x interfaces, M_s and H_k can be controlled independently through V_G -modulated O^{2-} distribution at the two interfaces.

In experiments, we show that all these states in Figs. 2(a)–2(e) can be reliably controlled in the sequence of $a \rightarrow b \rightarrow c \rightarrow d \rightarrow e \rightarrow b$, and the magnetism of the Co can be manipulated to any possible state. Moreover, two regions with a constant M_s around the pure metallic Co state [Fig. 2(c)] and a constant H_k during the $-V_G$ application [Fig. 2(d)] are obtained. These two special regions are further used for investigating SOT switching by altering either M_s or H_k independently.

C. Experimental demonstration of voltage-controlled metallic Pt/Co interfaces

The samples for demonstrating the voltage-control effects are the Si/SiO₂ (300 nm)/Pt (3 nm)/Co (0.9 nm)/GdO_x (80 nm) samples, which are patterned into Hall bar structures with a feature size of 2.5 μm . A schematic of the sample structure is shown as the inset of Fig. 3(d). V_G is applied at an elevated temperature between 200 – 260 $^\circ\text{C}$, while the magnetic states are detected at room temperature. Figures 3(a) and 3(b) show the typical vertical and in-plane field-dependent R_H , respectively. In Fig. 3(b), the H_x is tilted about 4 $^\circ$ toward the z direction to get a coherent magnetization switching. The M_s which is proportional to R_H [38] can be obtained from Fig. 3(a) and H_k can be evaluated from Fig. 3(b) through the Stoner-Wohlfarth model [39]. We first confirm that the magnetism could be controlled by V_G and then extract R_H and H_k as a function of V_G application time as shown in Fig. 3(c), where V_G is applied at 260 $^\circ\text{C}$. As shown in Fig. 3(c), from the as-deposited state ($R_H = 0.15 \Omega, H_k = 8.95 \text{ kOe}$), the sample is first set to a fully oxidized state with almost zero R_H and H_k under $-V_G$ application (first violet area on the left). Then, under $+V_G$ application, R_H monotonically increases to a maximum value of 0.48 Ω while H_k first increases and then decreases. The monotonical increase of R_H indicates that the CoO_x is continuously reduced to Co. The increase of H_k at the initial stage illustrates that the PMA is gradually established as O^{2-} is removed from the CoO film, and the subsequent decrease of H_k is due to a

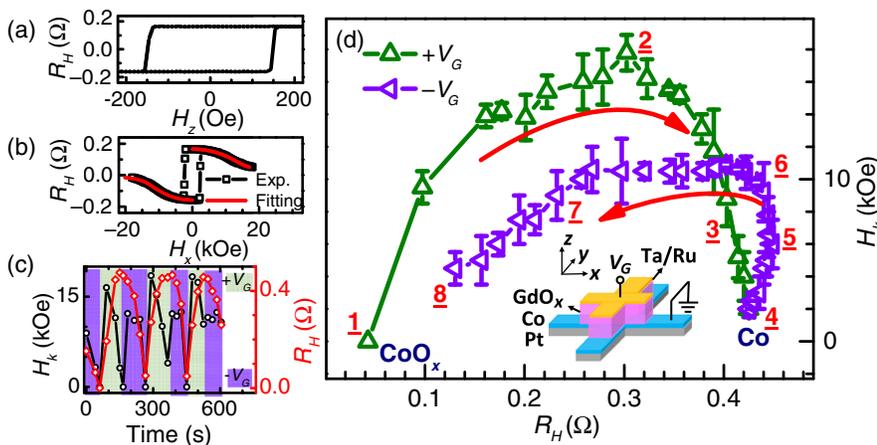


FIG. 3. The evolution of magnetism under V_G control. (a),(b) Typical R_H curves under vertical (a) or in-plane fields (b). The red lines in (b) are fitting results by using the Stoner-Wohlfarth model. (c) Typical time dependences of H_k and R_H under V_G control. (d) The measured H_k as a function of R_H in a typical V_G control cycle. The red arrows indicate the time evolution of the magnetism from states 1 to 8. Inset: the sample geometry and the experimental configuration of the V_G application.

thicker Co layer that prefers in-plane anisotropy [15]. Next, under $-V_G$ application, the sample returns to the PMA state and then to a fully oxidized state. The magnetic transition between the pure metallic state and the fully oxidized state can be reliably repeated, and all of these results are consistent with previous reports [15]. Note the magnetic states near the fully oxidized state, with zero R_H and H_k after $-V_G$ application, which are attributed to in-plane anisotropic states [16] probably correspond to superparamagnetic states [39].

We then inspect the detailed evolution of H_k and M_s with V_G applied at 200 °C. The lower temperature allows a finer control of H_k and M_s by voltage. In Fig. 3(d), we plot the extracted H_k as a function of R_H at each state after V_G application. As indicated by the red arrows, the V_G control cycle starts from an almost fully oxidized state (state 1) goes to a pure metallic state (state 4) under $+V_G$ application, and then back to another almost fully oxidized state (state 8) under $-V_G$ application. The remarkable feature of Fig. 3(d) is that the $H_k - R_H$ curve shows a hysteresislike behavior, where H_k has different values after $+V_G$ and $-V_G$ applications at the same R_H . Particularly, two regions are distinctive: from state 4 to 6, H_k sharply increases from 0 to about 10 kOe, while R_H keeps almost the same; from state 6 to 7, H_k keeps almost constant over a broad range of R_H .

The hysteresislike $H_k - R_H$ curve is significant because the same R_H , which corresponds to the same oxidation level of the Co layer and thus the same Co/oxide interface, should have the same H_k , if only the PMA at the Co/oxide interface is controlled by voltage. To explain these results, the Pt/Co interface must also be modified by the voltage through O^{2-} diffusion. According to Fig. 2, the states between 1 and 2 in Fig. 3(d) have a clean Pt/Co interface after $+V_G$ application, corresponding to Fig. 2(b) with a larger H_k , and the states between 7 and 8 have a contaminated Pt/Co interface after $-V_G$ application, corresponding to Fig. 2(d) with a weaker H_k . Different H_k values at the same R_H can be well understood in this way. This explanation is also directly confirmed by the STEM samples. The measured H_k of sample *f* in Fig. 1 is 4.3 kOe, larger than that of sample *d* ($H_k = 2.1$ kOe), even with the same R_H as shown in Fig. 1(b). Compared to sample *d*, sample *f* has a cleaner Pt/Co interface as shown in Figs. 1(d) and 1(f), well consistent with this explanation that the origin of larger H_k at the same R_H arises from a cleaner Pt/Co interface. Nevertheless, the H_k behaviors in the right minor loop 3-4-5-6 cannot be simply explained through Figs. 2(b) and 2(d). In this minor loop, H_k first quickly increases from 4 to 6 after $-V_G$ application and then reaches a saturation value while R_H keeps almost constant. After that, M_s decreases while H_k maintains nearly constant (6 to 7). Beyond 7, both M_s and H_k decrease when the Co film is gradually returned to CoO by $-V_G$. Contrary to the left major loop, H_k after $-V_G$

application in the minor loop is larger than that after $+V_G$ application at the same R_H .

To explain the minor loop, we consider the magnetic states starting from the metallic Co state 4 in which all the O^{2-} has been driven into the GdO_x . In the beginning of $-V_G$ application (4 to 5), the $-V_G$ -driven O^{2-} moves toward the Co/ GdO_x interface and bonds to the Co atoms to build PMA. Since the O^{2-} concentration is still very low at this initial state, the Co layer has not been significantly oxidized, consistent with the fact that R_H keeps almost constant and H_k increases as shown in Fig. 3(d). Additionally, the O^{2-} has not diffused into the Co layer and the Pt/Co interface keeps clean. Compared to the states between 3 and 4, the larger H_k between 4 and 5 are likely due to these O^{2-} at the Co/ GdO_x interface that contribute to the PMA but do not significantly oxidize the Co layer. For the states between 3 and 4, at the final stage of $+V_G$ application, all of the O^{2-} is depleted except those with strong a bond to Co in unreduced CoO_x , and thus there is no such O^{2-} only contributing to the PMA (all the O^{2-} strongly bonds to Co and contributes to the CoO_x for the states between 3 and 4) and H_k is smaller. From states 5 to 6, the increased rate of H_k change starts to decrease, although the H_k has not been saturated, and R_H begins to decrease. These magnetic states indicate that the O^{2-} has diffused to the Pt/Co interface and concentrated enough to start oxidizing the Co layer. The appearance of the minor loop further confirms the O^{2-} diffusion and the contribution from both the Pt/Co and Co/ GdO_x interfaces to the voltage control of PMA.

In principle, by controlling the O^{2-} at both the Pt/Co and Co/ GdO_x interfaces, all the possible magnetic states shown in the major and minor loops of Fig. 3(d) can be achieved. Figure 4(a) shows the controlled magnetic states starting from the same initial magnetic state with zero R_H and H_k , but the $-V_G$ are applied from different states. The controlled magnetic states where $-V_G$ are applied from $R_H = 0.33 \Omega$, 0.37Ω , and 0.41Ω are plotted in purple, red, and violet, respectively.

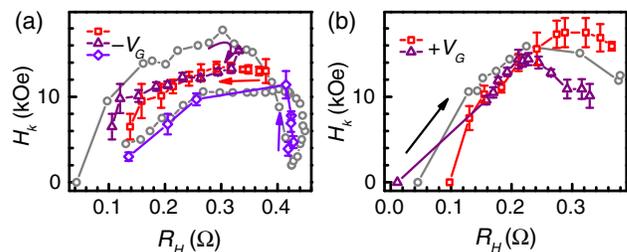


FIG. 4. The effects of reversal V_G applications on the control effects. (a) H_k versus R_H curves when $-V_G$ is applied from $R_H = 0.33 \Omega$ (purple), 0.37Ω (red), and 0.41Ω (violet). (b) H_k versus R_H curves where $+V_G$ is applied from different initial magnetic states. The gray curve is from Fig. 3(d) for comparison. The arrows in (a),(b) indicate the time evolution of controlled magnetic states.

red, and violet curves, respectively. The gray curve is from Fig. 3(d) for comparison. This figure shows that the states within the major and minor loops can be achieved by controlling the beginning of the $-V_G$ application. Figure 4(b) presents the magnetic states under $+V_G$ application, but starting from different initial states. These initial states are obtained through $-V_G$ applications. Thus, this plot presents the control effects that the beginning of $+V_G$ application leads to. It also shows that the controlled magnetic states depend on the beginning state where the reversal V_G is applied. Moreover, Fig. 4(b) shows a larger initial R_H value can result in a larger maximum H_k in the following control process. These results indicate that the magnetism can be controlled to any possible magnetic state by controlling the V_G application process.

D. Voltage control of SOT switching

SOTs offer a very promising way to switch magnetization electrically [11,40]. However, many aspects of SOT switching are still under debate. For example, the origin of SOTs can be spin Hall effects (SHEs) [40–42] or Rashba effects [11,43], and their roles during the switching process are still not clear [11,41,42,44–46]. Generally, the SOT switching starts from a reversal-domain nucleation in a uniform magnetization state and then the nucleated domains expand to the entire film [44,47,48]. Therefore, the critical SOT-switching current corresponds to the current for destabilizing the uniform magnetization, which is attributed to dampinglike [42,49] or fieldlike spin torques [45] as well as additional thermal effects [50–52]. Extracting the intrinsic magnetic property dependence of the SOT switching in the same sample will provide critical insights on the switching mechanism. However, it cannot be realized through a conventional method since the

intrinsic properties are specified during the sample fabrication. As demonstrated above, the M_s , H_k , as well as the oxidation level at both interfaces of the Pt/Co/GdO_x structures can be independently controlled by V_G , thereby offering a unique approach for studying SOT effects.

The voltage-controlled SOT switching is explored in samples with the structure of Si/SiO₂ (300 nm)/Pt (3 nm)/Co (0.6 nm)/GdO_x (80 nm) multilayers. The SOT switching is measured in the same way as previous studies [44,46,51]. Figure 5(a) shows typical current-driven SOT-switching curves under in-plane fields. The magnetization switching direction depends on both the current and field directions, which is consistent with previous reports [11,40,41,44,46–48,51]. We choose the current where R_H begins to change as I_c , as marked in Fig. 5(a). As mentioned above, I_c corresponds to the critical current for destabilizing a uniform magnetization, under which the reversal domains start to nucleate. Figure 5(b) shows the switching curves after $-V_G$ and then $+V_G$ applications. One can see that I_c can be greatly decreased by $-V_G$ together with the reduced M_s , and increased by $+V_G$ in the opposite fashion.

Figure 5(c) shows the evolution of H_k and R_H under the $\pm V_G$ control. The H_k - R_H curve exhibits a similar hysteretic behavior as in Fig. 3(d). During the application of $-V_G$, H_k can be maintained nearly constant at 5 kOe over a large R_H region (0.15–0.35 Ω , states between 6 and 7). A minor loop also exists near the pure Co state, where a large change of H_k (from 2 to 5 kOe) is accompanied by a small change of R_H (states between 4 and 5). I_c is plotted against R_H in Fig. 5(d). Remarkably, I_c can be reduced by more than 10 times with the application of $-V_G$, demonstrating the advantage of controlling SOT switching by ionic effects. The modulation of I_c is much larger than that by using an ionic liquid gate [53] in which only the FM-oxide interface is controlled. Particularly, in the two special regions, it is

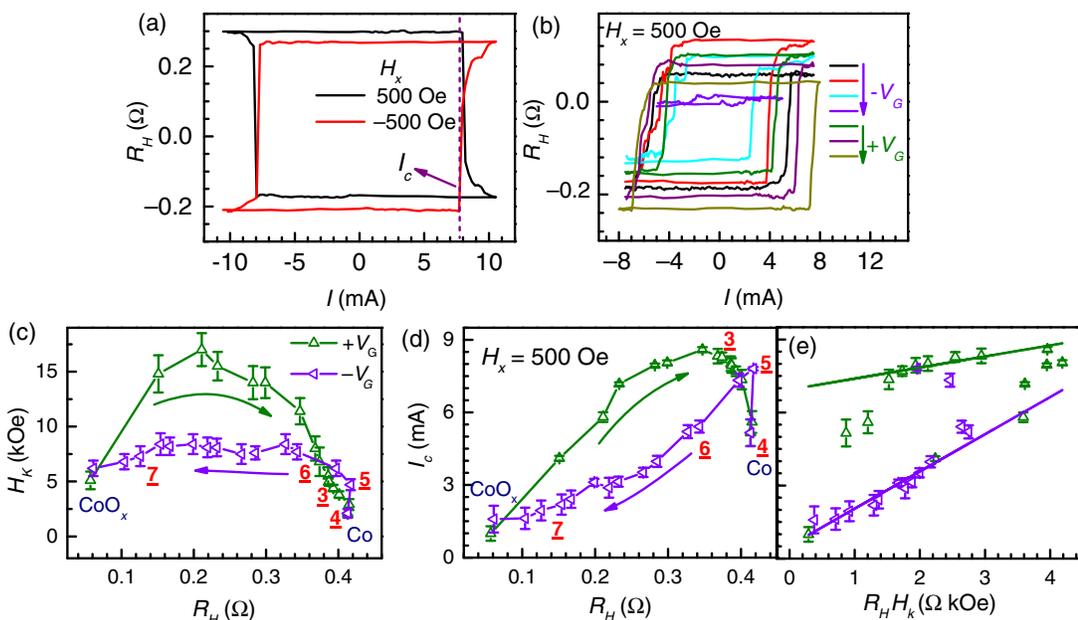


FIG. 5. SOT switching under V_G control. (a) Typical current induced SOT switching under in-plane fields. (b) The SOT-switching curves under V_G control. (c), (d) The extracted H_k and corresponding I_c (d) as a function of R_H at each magnetic state controlled by V_G . The arrows indicate the time evolution of control process. (e) I_c is replotted as a function of $R_H H_k$. The solid lines are the linear fitting results.

clearly shown that I_c is proportional to R_H (states 6 and 7) and also strongly depends on H_k (states 4 and 5).

To analyze the SOTs and their effects on the magnetization switching, we adopt a simple magnetic switching model by only considering dampinglike torque [42,50]. Although the fieldlike torques [45] and possible DMI [54] are not counted, this simple model can explain several experimental observations [55,56], and thus may provide some initial insights into our results. According to this model [42,50],

$$I_c = \frac{2e M_s t_F}{\hbar \theta_{SH}} \left(\frac{H_k}{2} - \frac{H_x}{\sqrt{2}} \right), \quad (1)$$

where θ_{SH} is the effective spin Hall angle and t_F is the thickness of ferromagnetic layer. By considering $R_H \propto M_s t_F$ and $H_x \ll H_k$,

$$I_c \propto \frac{R_H H_k}{\theta_{SH}}, \quad (2)$$

where the slope of I_c versus $R_H H_k$ represents the θ_{SH} in this system. Figure 5(e) shows the measured I_c as a function of $R_H H_k$ by replotting Figs. 5(c) and 5(d), and the solid lines are the linear fitting results. Note the data points obtained after the application of $+V_G$ and $-V_G$ are plotted as up triangles and left triangles, respectively. Immediately, two linear regions can be identified, generally confirming the validity of this simple model. The first linear region is in the range of $I_c < 5.5$ mA, in which the slopes of I_c versus $R_H H_k$ is about 1.57×10^7 A cm⁻² Ω⁻¹ kOe⁻¹. Another linear region is within the range of 7.0 mA $< I_c < 9.0$ mA. This region corresponds to the states with a clean Pt/Co interface after $+V_G$ application. The slope in this region is about 0.47×10^7 A cm⁻² Ω⁻¹ kOe⁻¹. This slope corresponds to a 3.5 times larger θ_{SH} than that after $-V_G$ application, demonstrating that a clean Pt/Co interface plays a critical role in the efficiency of SOTs. Note the Joule-heating-induced thermal effects during the magnetization switching cannot explain the larger θ_{SH} after $+V_G$ application [39].

IV. CONCLUSIONS

HM-FM interfaces play a critical role in many interfacial phenomena, such as spin Seebeck effects [29], spin Hall magnetoresistance [30], and other magnetic proximity effects [31]. We have demonstrated that the metallic HM-FM interface can also be controlled electrically, providing opportunities for electrical modulation of these interfacial effects. Contrary to previous studies where only the FM-oxide interface was controlled and the effect was limited to a few special states, the controlled HM-FM interface strongly broadens the magnetic states in the field of voltage-controlled magnetism. In addition, M_s and H_k can be manipulated independently and the magnetism can be set to any possible magnetic state. STEM and EELS

results directly confirm the controllable metallic interface and its effects on the control of H_k at the same R_H .

Recently, it has been shown that the PMA in a HM-FM-metal-oxide structure strongly depends on the thickness of the top capping metal layer [16,36,57]. The speculated reasons focused only on the influence of the capping metal layer on the FM-metal interface. This work indicates an alternative mechanism that the thickness of the capping layer may alter the O⁻² distribution underneath at both the HM-FM and FM-metal interface during the sputtering process. By considering the contributions to the PMA from both interfaces, the change of the O⁻² distribution at the two interfaces may result in the capping metal layer dependence on the PMA.

Furthermore, we show that the efficiency of SOTs can also be controlled and the critical SOT-switching current can be reduced about 10 times. Previous studies have indicated that the efficiency of spin injection at the interface depends on the transparency of this interface [27,28], but demonstrated in different samples that inevitably also alter other conditions. Our results clearly show that the effective SOTs can be enhanced about 3.5 times with a clear HM-FM interface in the same sample. The unique voltage control of magnetism may also be applied to investigate other magnetism-related effects by modulating the M_s and H_k independently in the same sample, avoiding the influence of the variation between different samples in conventional measurements.

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Note added in proof.—We noticed a recent work by Liu *et al.* reporting control of nonlinear spin-oscillation by electrostatic gating [58], which was attributed to the VCMA effect and voltage-controlled SOTs originated from FM-oxide interfaces [59].

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