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ABSTRACT

The evolution of skyrmion crystals encapsulates skyrmion’s critical behaviors, such as nucleation, deformation, and annihilation. Here, we achieve a tunable evolution of artificial skyrmion crystals in nanostructured synthetic antiferromagnet multilayers, which are composed of perpendicular magnetic multilayers and nanopatterned arrays of magnetic nanodots. The out-of-plane magnetization hysteresis loops and first-order reversal curves show that the nucleation and annihilation of the artificial skyrmion can be controlled by tuning the diameter of and spacing between the nanodots. Moreover, when the bottom layer thickness increases, the annihilation of skyrmion shifts from evolving into a ferromagnetic spin texture to evolving into an antiferromagnetic spin texture. Most significantly, nonvolatile multiple states are realized at zero magnetic field via controlling the proportion of the annihilated skyrmions in the skyrmion crystal. Our results demonstrate the tunability and flexibility of the artificial skyrmion platform, providing a promising route to achieve skyrmion-based multistate devices, such as neuromorphic spintronic devices.

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I. INTRODUCTION

A magnetic skyrmion is a topologically protected spin texture with nontrivial spin configurations.1–4 It was first observed in noncentrosymmetric B20 compound MnSi at a low temperature, about 29 K,5 and was stabilized by an external magnetic field and bulk Dzyaloshinskii–Moriya interaction (DMI).6–8 It was found that other bulk magnets with non-centrosymmetry and DMI can also host skyrmion configurations.9–11 Alternatively, the existence of skyrmions in nanometer-thick multilayers with interfacial DMI was also theoretically predicted9,12 and experimentally observed in the Ta/CoFeB/TaOx multilayer.17 Since then, magnetic multilayers consisting of ferromagnets and heavy metals become promising and attractive skyrmion platforms.7,12 In particular, the room-temperature yet zero-field skyrmion can nucleate and exist stably in this platform of multilayers, facilitating the development of practical skyrmion-based spintronic devices.7,12–20

Because of their topological stability, nanoscale size, and low driven current, magnetic skyrmions are highly promising for both fundamental studies and spintronic applications,21–23 such as racetrack memory,24,25 logic devices,26,27 and skyrmion magonic crystals.17,28,29 Incorporating skyrmions into advanced artificial structures, such as nanoribbons,30–43 nanowires,44,45 nanodots,46,47 and hybrid nanodot-film structures,48–53 is an emerging trend in the investigation of the skyrmionic field. Although the nano-patterns inevitably limit their motion along racetracks resulting from the pinning effect, skyrmions can in turn locally oscillate in nanodots.54–55 Most importantly, in contrast to the spontaneous nucleation in bulk materials and continuous magnetic multilayers, skyrmions in nano-patterned multilayers exhibit higher existence stability, position controllability, and even a crystal order. The crystal-like group formed by these periodic skyrmions is referred to as the skyrmion lattice50,56 or skyrmion crystal.48,56,57 Recently, the idea of endowing the skyrmion array with an artificial crystal order attracts much research attention and expands...
the spintronic research frontier. However, to enhance and exploit the functionalities of the skyrmion crystals, introducing engineered spin textures into the skyrmion crystals is desirable yet elusive.

To construct a skyrmion crystal, there are typically two types of skyrmions, namely, conventional skyrmion induced by DMI and artificial skyrmion induced by interlayer interaction. In particular, the artificial skyrmion crystal gains more functionalities due to the extra tunability from the interface. Previous works have demonstrated that artificial skyrmion crystals can be formed via the interfacial ferromagnetic coupling between the top magnetic nanodots and the bottom multilayer with perpendicular magnetic anisotropy, or via the exchange bias in the ferromagnet/antiferromagnet multilayers. Recently, antiferromagnetic interlayer exchange coupling (AFM-IEC) in a synthetic antiferromagnet (SAF) has drawn much attention due to its capabilities in stabilizing skyrmions or generating functional spin textures. Therefore, engineering the artificial skyrmion crystal in SAF nanostructures might introduce controlled spin textures, permit unseen functionalities, and boost the control of skyrmions.

In this work, we engineer the SAF into nanostructures, which contain the top nanodots and the continuous bottom layer with varied geometries. A periodically ordered skyrmion crystal is generated in these nanostructures. The out-of-plane magnetization hysteresis loops and first-order reversal curves (FORCs) reveal that the evolutions of the skyrmion crystal, which are nucleation and annihilation, strongly depend on the thicknesses of the bottom magnetic layer and the diameters of the top nanodots. Especially, by increasing the thickness of the bottom layer, skyrmions can be switched from evolving into a ferromagnetic to an antiferromagnetic spin texture. We also find that even with part of the skyrmion crystal evolving into other spin textures, the rest is still stable. Based on this feature, we achieve nonvolatile multiple states, characterized by multiple levels of Hall resistance.

II. RESULTS AND DISCUSSION

Figure 1(a) shows the schematic structure of the nanostructured SAF multilayers Ta(4)/Pt(4)/[Pt(0.6)/Co(0.6)]N/Ru(0.9)/[Co(0.6)/Pt(0.6)]4/Ta(4), where the numbers in parentheses are the nominal layer thickness in nanometer. Detailed information of sample preparation can be found in Sec. IV. The spacing layer Ru provides an AFM-IEC between the bottom [Pt/Co]N and top [Co/Pt]4 layers. The top Ta layer is used to prevent oxidation, and the bottom Ta and Pt layers work as seeding layers. The bottom [Pt/Co]N layer has a repeating number N (N = 2 or 4). Note that the repeating number N = 2 or 4 ensures a single-magnetic-domain state in the bottom [Pt/Co]N layer. When N takes a larger value, the domain in the magnetic layer may exhibit a multi-domain state, which is unfavorable in this work. The top [Co/Pt]4 layer was etched into circular nanodots with varying diameters.

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diameters ($d$) varying from 100 to 600 nm. In all samples, the spacing between the nanodots (S) is equal to $d$. We prepared the nanostructured SAF multilayers with two types of nanodot distributions, namely, (i) two-dimensional arrangement of nanodots for magneto-optical Kerr effect (MOKE) measurement [see the scanning electron microscopy (SEM) image of $d = 600$ nm at the bottom left of Fig. 1(a)] and (ii) a finite number of nanodots (Q) for Hall resistance measurement [see the SEM image of $d = 400$ nm at the bottom right of Fig. 1(a) for $Q = 9$]. Note that $Q = 0$ is the sample with continuous ferromagnetic multilayers [Pt/Co]$_2$, and it serves as a reference sample. The schematics of the samples for MOKE and Hall resistance measurements can be found in Sec. II of the supplementary material. We define the upward and downward directions as the positive and negative directions of the external magnetic field, respectively [Fig. 1(a)].

The $m_r$–$H$ loops of nanostructured SAF multilayers were measured by MOKE. Technical details of the MOKE measurement are presented in Sec. IV. Figures 1(b) and 1(c) show the descending branches of the MOKE-measured $m_r$–$H$ loops of the nanostructured SAF multilayers with $d = 600$ nm for $N = 2$ and $N = 4$, respectively. Both of the $m_r$–$H$ loops exhibit multi-step switching, which corresponds to different magnetization distributions. We define the coercive field ($H_C$) as the magnetic field value at which the $m_r$ crosses zero (see Sec. III of the supplementary material). Additionally, to identify all the magnetization states, we performed magnetic force microscopy (MFM) measurements on the nanostructured SAF multilayers with $N = 2$, $d = 600$ nm at various fields. Figure 1(d) shows the measured typical MFM images at selected fields ($H = 12$, 0.8, $-2.6$, and $-12$ kOe) for different magnetization states. Technical details of the MFM measurement are presented in Sec. IV. We found that, along the descending branch of the $m_r$–$H$ loop, the nanostructure with $N = 2/4$ and $d = 600$ nm successively experiences four/five distinct magnetization states. In order to elaborate the different magnetic transitions clearly for $N = 2/4$, we presented six examples of cross section views of magnetization distribution at different magnetization states, which are labeled by numbers from 1 to 6 on the $m_r$–$H$ loops [insets of Figs. 1(b) and 1(c)].

We now focus on the process of magnetization switching of the $N = 2$, $d = 600$ nm samples [Fig. 1(b)]. When the external field is larger than the saturation field ($H_{Sat}$), the magnetization of the top and bottom layers is aligned along the direction of the external field [inset 1 of Fig. 1(b)]. As the external field decreases toward 0 kOe, the magnetization of the top nanodots is reversed by the AFM-IEC [inset 2 of Fig. 1(b)]. Because the magnetizations of the top nanodots and the bottom continuous layer are antiparallel, we denote this magnetization configuration as an antiferromagnetic spin texture (AST) state. When the negative external field is greater than the coercive field ($H_C$), the magnetization of the dot-uncovered region of the bottom layer is reversed downward. The magnetization of the dot-covered region remains antiparallel to that of the top nanodots because of the AFM-IEC. Considering solely the bottom continuous layer, a skyrmion-like magnetization configuration nucleates [inset 3 of Fig. 1(b)], which is referred to as a skyrmion (SK) state. Further increasing the external field negatively will gradually offset the AFM-IEC protection, resulting in a shrinking annihilation of the skyrmion [inset 4 of Fig. 1(b)]. Because the negative external field is larger than $H_{Sat}$, the magnetization of the bottom layer is fully reversed downward, and this state is referred to as a ferromagnetic spin texture (FST) state [inset 5 of Fig. 1(b)]. The MFM images shown in Fig. 1(d) reflect the magnetization configurations corresponding to the four representative states of the $m_r$–$H$ loop in Fig. 1(b): (i) positive FST state ($H = 12$ kOe); (ii) AST state ($H = 0.8$ kOe); (iii) SK state ($H = -2.6$ kOe); and (iv) negative FST state ($H = -12$ kOe).

Similarly, Fig. 1(c) shows the magnetization evolution of nanostructured SAF multilayers with $N = 4$, $d = 600$ nm. In particular, an additional AST state with a rise of $m_r$ occurs after the annihilation of the artificial skyrmions [inset 6 of Fig. 1(c)]. In comparison to the $N = 2$ nanostructure, the additional AST state is induced by a larger AFM-IEC in the $N = 4$ nanostructure. The difference of AFM-IEC between the $N = 2$ and $N = 4$ nanostructures is confirmed by comparing the $m_r$–$H$ loops of continuous SAF multilayers with different $N$ (Sec. IV of the supplementary material).

In order to investigate the effect of $d$ on the nucleation and shrinking annihilation of skyrmions in nanostructured SAF multilayers, $d$ was varied from 100 to 600 nm. Figures 2(a) and 2(b) show the fractions of the MOKE-measured $m_r$–$H$ loops with various $N$ and $d$ as well as the reference samples with continuous ferromagnetic [Pt/Co]$_2$ and [Pt/Co]$_4$ multilayers. The full $m_r$–$H$ loops are shown in Sec. V of the supplementary material. The critical external field of skyrmion nucleation ($H_C$) decreases and gradually approaches the $H_z$ of the reference samples, when $d$ increases from 100 to 600 nm for both $N = 2$ and 4 samples. On the contrary, the critical external fields of skyrmion shrinking annihilation, which are $H_{Sat}$ for $N = 2$ and $H_{AST}$ for $N = 4$, increase as $d$ decreases. The different dependence of $H_C$, $H_{Sat}$ and $H_{AST}$ against $d$ can be understood from the different switching scenarios of the bottom continuous [Pt/Co]$_N$ layer. Taking $N = 2$ as an example, Fig. 2(c) schematically shows that the nanostructured SAF multilayers can be divided into three regions, namely, (a) the dot-covered region, (b) the dot-non-covered region, and (c) the top nanodots. When the external field negatively increases, the magnetization transitions of the AST-SK and the SK-FST manifest as the switching of regions B and A, respectively [Fig. 2(d)]. The AST-SK magnetization transition is realized by the reversal of region B, which is collectively determined by three interactions: (i) the coercivity field ($H_{c,B}$) of the continuous bottom [Pt/Co]$_2$ layer, (ii) the ferromagnetic exchange interaction field ($H_{ex-AB}$) at the interface between regions A and B, and (iii) the external field ($H$). Figure 2(d) shows that, during the skyrmion nucleation, the magnetization of region B is reversed when $H$ is larger than the AST-SK transition field ($H_z$). The value of $H_z$ is decided by the summation of $H_{ex-AB}$ and $H_{c,B}$.

$$H_z = |H_{ex-AB}| + |H_{c,B}|,$$

(1)

where both $H_{ex-AB}$ and $H_{c,B}$ protect the region B from being reversed by $H$. However, when $H = H_z$, the magnetization of region B reverse downward, the magnetization of region A is retained by the AFM-IEC between regions A and C. As a result, region A tends to protect the region B from being reversed by $H_{ex-AB}$. For a smaller $d$ ($d = S$), the contribution of $H_{ex-AB}$ to $H_z$ is relatively larger. While $H_{c,B}$ is dependent on the coercivity field of the continuous [Pt/Co]$_2$ layer but independent of $d$, the inverse dependence of $H_z$ on $d$, therefore, can be understood according to Eq. (1).

On the other hand, the SK-FST magnetization transition is realized by the reversal of region A. This transition is collectively determined by four parameters: (i) the coercivity field ($H_{c,A}$) of the continuous bottom [Pt/Co]$_2$ layer, (ii) the ferromagnetic exchange interaction field ($H_{ex-BA}$) at the interface between regions A and B, (iii)
Therefore, the field ranges of the SK state, which are considering the field-dependent reversal of regions A, B, and C.

However, both are independent of $H$ where both $H_{ex-CA}$ and $H_{ex-BA}$ protect region A from being reversed by $H$. $H_{ex-CA}$ and $H_{ex-BA}$ are determined by the AFM-IEC strength and the coercivity field of the continuous $[\text{Pt/Co}]_2$ layer, respectively. However, both are independent of $d$. In contrast, region B tends to reverse region A via $H_{ex-BA}$. For a smaller $d$, the $H_{ex-BA}$ is relatively larger. Hence, the dependence of $H_{sat}$ on $d$ can be understood according to Eq. (2), and a smaller $d$ results in a smaller $H_{sat}$. Similarly, the dependence of $H_{c1}$ and $H_{AST}$ on $d$ for $N=4$ can also be understood by considering the field-dependent reversal of regions A, B, and C. Therefore, the field ranges of the SK state, which are $|H_{sat} - H_{c1}|$ for $N=2$ and $|H_{AST} - H_{c1}|$ for $N=4$, can be manipulated by tailoring $d$, and the tunability further enables the flexibility of the nanostructured SAF multilayers in skymionic and magnonic applications.

Furthermore, in contrast to the skymion’s shrinking annihilation in the form of SK-FST and SK-AST transitions, skymions can also annihilate in a form of expansion. To investigate the skymion-expansion-induced magnetization transition, we measured minor $m_z$-$H$ loops using MOKE, i.e., $H_{sat}$-$H_R$-$H_{sat}$ loops, by choosing a reversal field $H_R < H_{sat}$. Figures 3(a) and 3(b) show minor loops of $d=300$ nm samples with $N=2$ and 4, respectively. Each figure contains two minor loops with $H_R$ located in the SK state ($H_R = H_{R1}$) and FST/AST state ($H_R = H_{R2}$). We observed that the magnetization $m_z(H, H_R)$ is highly dependent on the selection of $H_R$, when the $H$ increases from $H_R$ toward $H_{sat}$. For instance, the corresponding cross sections of three loops with different $H_R$ at $H = 2$ kOe are indicated by vertical dashed lines in Figs. 3(a) and 3(b). This difference in magnetization distribution originates from the $H_R$-dependent magnetization transition. When $H_R = H_{R1}$, the SK-AST transition occurs in both $N=2$ and 4 samples together with a type of skymion annihilation in a form of expansion.

Because the skymion-expansion-induced magnetization transition depends on the value of $H_{R1}$, a systematic investigation of $m_z(H, H_R)$ is necessary. Here, we used the FORC technique for the detailed exploration. The FORC technique was used in the MOKE measurement by changing $H_R$ with an interval of $\Delta H_R$. The FORCs, which is composed of a large number of “$H_{sat}$-$H_R$-$H_{sat}$” minor loops, have been widely used to study the distribution of switching fields, the interaction fields between neighboring domains, and the irreversibility of the magnetization switching mechanism. Details of FORCs and calculation of the FORC diagram are described in Sec. VI of the supplementary material. Figures 3(c) and 3(d) show the fractions of MOKE-measured FORCs in the form of contour plots of $m_z(H, H_R)$ with $d$ varying from 100 to 600 nm for $N=2$ and 4 samples. The corresponding full FORCs and FORC diagrams can be found in Fig. S6 of the supplementary material. Depending on the values of $H_{R1}$, there are two left-right boundaries in the contour plots of $m_z(H, H_R)$, and they are indicated as boundaries 1 and 2 in Figs. 3(c) and 3(d). Now, we use the minor loops and the contour plots of $m_z(H, H_R)$ for $N=2$, $d=300$ nm [Figs. 3(a) and 3(c)] to elaborate on the origins behind the presence of boundaries. When $H_R = H_{R1}$, boundary #1 corresponds to the SK-AST transition [ insets of Fig. 3(c)]. As for $H_R = H_{R2}$, boundary #2 corresponds to the AST-SK transition [ insets of Fig. 3(c)]. In addition, the position of boundary #1/#2 decreases/increases to low/high $H$ when $d$ decreases from 600 to 100 nm. This dependence of boundary position indicates that the skymions in larger $d$ samples can expand continuously to the positive field region. On one hand, skymion can stabilize under zero field, when $H_R$ locates at the SK state. On the other hand, the skymions can nucleate more easily when $H_R$ locates at the FST state for the samples with larger $d$.

Figure 3(d) shows the contour plots of $m_z(H, H_R)$ for $N=4$ sample with two left-right boundaries. Similar to the boundaries of $N=2$
sample, the boundary #1/#2 corresponds to the SK-AST and the AST-SK transition for \( N = 4 \) sample by considering the minor loops in Fig. 3(b). The boundary #1/#2 of the \( N = 4 \) sample moves similarly to that of the \( N = 2 \) sample. Note that no boundary #1 occurs in \( N = 4 \) and \( d = 100 \) nm sample, indicating that an SK state hardly exists in this case. This is consistent with our experimental results of the \( m_z-H \) loop of \( d = 100 \) nm in Fig. 2(b).

So far, we have mainly focused on the intrinsic properties of skyrmions in the nanostructured SAF multilayers with various \( N \) and \( d \). Now, we switch to Hall resistance measurements to investigate the potential applications based on the evolution of skyrmions. Figure 4(a) shows the evolution of the Hall resistance \( (R_{xy}) \) of the nanostructured SAF multilayers with fixed \( N = 4 \) and \( d = 400 \) nm, but varied \( Q = 1 \) and 9. In order to reveal the influence of \( Q \) on the behaviors of skyrmion, a zoomed-in view of the SK region is presented in Fig. 4(b). The SK region can be divided into stabilization and annihilation regions, which correspond to the shrinking of the skyrmion and the SK-AST transition, respectively. We found that the annihilation region of skyrmions in \( Q = 9 \) nanostructure is much larger than that in the \( Q = 1 \) nanostructure, which indicates that the nine skyrmions in the \( Q = 9 \) nanostructure annihilate asynchronously. Based on this phenomenon, we could purposely nucleate or annihilate different numbers of skyrmions to achieve multiple states at zero field. Note that the annihilation region manifests as a rise in the \( m_z-H \) loop [Fig. 1(c)] but a sharp drop in the Hall resistance hysteresis \( (R_{xy-H}) \) loop. This should be because Hall resistance measurement is more sensitive to the continuous bottom layer and less sensitive to the etched top layer.

To experimentally realize the multiple states based on the different numbers of skyrmions, we measured the reversal curves with the reversal fields \( H_R \) selected at different fields, where there are different numbers of skyrmions. Figure 4(c) shows a full major curve and three reversal curves with \( H_R = 1.6 \), \( 2.2 \), and \( 2.4 \) kOe. Figure 4(d) shows the zoomed-in view of the three reversal curves. Note that we only present the results of the descending branch. Similar states could also be achieved in the ascending branch of the positive field region.

As shown in Fig. 4(d), we obtained four different states at zero field, namely, base state, state 1, state 2, and state 3. The base state was realized by sweeping the external field from \( H_R = H_{sat} \) to zero field, and there is no skyrmion at the \( H_R \). State 1, 2, and 3 were achieved by sweeping the external field from different \( H_R \) to zero field, and there are different numbers of skyrmions at these selected \( H_R \). Moreover, we conducted local hysteresis loops sweeping between \( H_R \) and zero field for the four states. The value of \( R_{xy} \) at zero field is always repeatable.
which indicates that the multiple states based on the different numbers of skyrmions are nonvolatile and robust. We attribute these multiple states to an asynchronous evolution of the skyrmions with the external field. For the nanostructured SAF multilayers with the periodically arranged nanodots, the skyrmions forms in the bottom layer are also arranged periodically and behave the same as their neighbors. However, the nine skyrmions in the nanostructured SAF multilayers with \(Q = 9\) can be divided into three types according to the difference in their positions: (i) four skyrmions at the corner, (ii) four skyrmions at the edge center, and (iii) one skyrmion at the center. The evolutions of the three types of skyrmions with the external field are different, which is confirmed by the micromagnetic simulation in Sec. VIII of the supplementary material (the relevant simulation parameters are given in Sec. I of the supplementary material). Our Hall resistance measurement and simulated results further demonstrate that, if we could introduce additional asymmetry of skyrmions through tailoring the shape, distribution, and size of the nanostructure, the nucleation and annihilation of skyrmion can be controlled more accurately.

III. CONCLUSION AND OUTLOOK

In summary, we experimentally investigated the behaviors of skyrmion crystals in the nanostructured SAF multilayers and validated a feasible scheme for the application of the multilayers. The existence of the skyrmion crystal in nanostructured SAF multilayers and its shrinking annihilation, i.e., the SK-FST and SK-AST transitions, were characterized by the MOKE-measured \(m_y-H\) loops. We also proved that the external field range of the existence of skyrmions could be improved via increasing the diameter of top nanodots \(d\). Moreover, with the help of the FORC technique, we found that the zero-field stability and field range of existence of skyrmions can be improved by choosing the reversal field at the SK region and tailoring \(d\). Finally, according to the asynchronous annihilation of skyrmions, we achieved the zero-field nonvolatile multiple states.

Since its discovery, magnetic skyrmion has been proposed for racetrack memory \(^{16}\) and logic gate \(^{35}\) based on the current-driven "continuous" motion along nanowires. Alternatively, the current-driven "local" motion or oscillation of skyrmion in nanodots has also
been demonstrated and proposed as microwave nano-oscillators.\textsuperscript{47,53,55} Even an electrically configurable pixelated skyrmions on nanoscale magnetic grids has been numerically demonstrated.\textsuperscript{9,51} With the application of electric current, our proposed skyrmions could oscillate in the dots-covered region in the continuous bottom layer and can be used for skyrmion-based microwave nano-oscillators.\textsuperscript{47,53,55} In addition, our reported skyrmion states can also be used to produce the zero-field multiple states with memristive behaviors. Hence, both the oscillatory motion and the multistate behavior of our proposed skyrmions could be used for developing future spintronic neuromorphic devices,\textsuperscript{92} including nano-oscillator-based magnetic neurons\textsuperscript{93,94} and spin-texture-based synapse.\textsuperscript{95–97}

**IV. METHODS**

**A. Sample preparation**

The used synthetic antiferromagnetic multilayers in this work, $\text{Ta(4)/Pt(4)/[Pt(0.6)/Co(0.6)]$_N$/Ru(0.9)/[Co(0.6)/Pt(0.6)]$_N$/Ta(4)}$, were sequentially deposited over oxidized silicon wafers using a DC magnetron sputtering technique. $N$ is 2 or 4 in our work. The numbers in the bracket are the nominal thickness of the corresponding layer. During the deposition, $2.3 \times 10^{-3}$Torr Argon gas was filled in the vacuum chamber with a base pressure of $2 \times 10^{-8}$Torr. The deposition rates were 0.21, 0.14, and 0.10 Å/s for Co, Pt, and Ru, respectively. The multilayers were then spin-coated with polymethyl methacrylate (PMMA). Electron beam lithography (EBL) was then used to pattern the PMMA layer with circular nanodots. The diameter and spacing of the circular nanodots varied from 100 to 600 nm. Then, $\text{Ar^+}$ ion milling was used to remove the part of the sample above Ru but uncovered by circular PMMA nanodots. Before etching the sputtered SAF multilayers, the etching rates of Ta, Pt, Co, and Ru were separately calibrated and optimized in the AIA ion milling system. During the etching process, the milling process is immediately stopped at a time of $t_M$ when Ru is detected with the help of the element monitor function of the ion milling system. For each $t_M$, several reference samples were prepared to affirm that the presented samples with the top $[\text{Co/Pt}]_N$ layer were fully etched and the bottom $[\text{Pt/Co}]_N$ layer was not etched. Finally, we dissolved the remaining PMMA on the sample and obtained patterned nanodots above Ru. The steps above are used to prepare the samples for MOKE measurements. To prepare the samples for Hall resistance measurements, we further etched the Ru layer and the bottom $[\text{Pt/Co}]_N$ layer below Ru into a crossbar shape (length 40 μm and width 3 μm) using EBL and $\text{Ar^+}$ ion milling methods. The samples used for MOKE measurements have periodic nanodots, while the samples used for Hall resistance measurements have a single nanodot or nine nanodots as shown in Fig. 1(a).

**B. Magneto-optical Kerr effect and Hall resistance measurements**

The $m_z$–$H$ loops and the FORCs were measured by the MagVision Kerr imaging system. Differential imaging technology and piezoelectric actuators were applied to enhance the detected magnetic signal and to eliminate the influence of sample drift, respectively. Refer to https://www.vertisis.com.sg/. The Hall resistance measurements were conducted with an AC/DC source of Keithley 6221 and a nanovoltmeter of Keithley 2182A. The current used in all the Hall resistance measurements was a square wave with a peak-to-peak amplitude of 0.2 mA.

Magnetic Force Microscopy; MFM measurements were conducted in a home-built system. This system has a superconducting magnet with a maximum magnetic field of 20 T.\textsuperscript{100,101} The piezoresistive cantilever of the system is the commercial PRC400 from Hitachi High-Tech Science Corporation. This cantilever has a 42 kHz resonant frequency. The MFM tip is coated with Cr, Fe, and Au films. The thicknesses of Cr, Fe, and Au coating are 5, 50, and 5 nm, respectively. The coercivity and saturation fields of the MFM tip are $H_c \approx 0.25$ kOe and $H_{sat} \approx 2$ kOe, respectively. The MFM uses a built-in phase-locked loop to adjust the scanning process and process signals. During the MFM measurement, the contact mode was first used to obtain a topographic image. According to the topographic image, the sample surface tilting along the fast and slow scan axes can be compensated. Second, the MFM images were measured in a frequency-modulation mode with a tip height of $\sim 100$ nm.

**SUPPLEMENTARY MATERIAL**

See the supplementary material for additional experimental and simulated results: micromagnetic simulation details; schematics of the samples for MOKE and Hall resistance measurements; definition of the coercive field; strength of antiferromagnetic interlayer exchange coupling; full descending branches of $m_z$–$H$ loops of nanostructured SAF multilayers; FORC measurement; analysis of the FORC diagram; and asynchronous evolution of skyrmions.

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**AUTHOR DECLARATIONS**

**Conflict of Interest**

The authors have no conflicts to disclose.

**Author Contributions**

M.M. and K.H. contributed equally to this work.

**DATA AVAILABILITY**

The data that support the findings of this study are available from the corresponding author upon reasonable request.
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