

Magnetic reversal behavior and domain morphology in low-damping CoFeB/Ni multilayers

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Perpendicularly magnetized multilayers (MLs) with low magnetic damping and highly periodic stripe domains are promising materials for energy-efficient spintronic and magnonic devices. Here, we investigate the structure, together with the static and dynamic magnetic properties of a polycrystalline sputter-deposited [CoFeB(0.3 nm)/Ni(0.6 nm)]₅₀ ML and compare it to a [Co(0.3 nm)/Ni(0.6 nm)]₅₀ reference stack. Scanning transmission electron microscopy, supplemented by atomic force microscopy, reveals that using a CoFeB/Ni ML yields a more cohesive and denser grain microstructure than in the Co/Ni ML. Our structural results are corroborated by reciprocal-space x-ray reflectivity and x-ray diffraction, further indicating a smoother CoFeB/Ni ML with weakened crystallographic texture and reduced crystalline coherence lengths compared to the Co/Ni ML, while preserving the overall fcc(1 1 1) growth induced by the Pt seed layer. Vibrating sample magnetometry shows that the CoFeB/Ni ML exhibits weak perpendicular magnetic anisotropy (PMA), resulting in near-zero-remnance isotropic labyrinth stripe domains following out-of-plane saturation and parallel stripe domains after in-plane saturation, consistent with the absence of strong pinning sites. In contrast, the Co/Ni ML exhibits stronger PMA and more pinning-mediated, disordered domain patterns due to its more inhomogeneous microstructure. Field-dependent magnetic force microscopy demonstrates an evolution from labyrinth stripe to dense bubble domain states in the CoFeB/Ni ML, accompanied by a reduction in microscopic return-point memory of the remanent state. Broadband ferromagnetic resonance measurements further disclose a 25% reduced Gilbert damping parameter α in the CoFeB/Ni ($\alpha = 0.0147$) compared to the Co/Ni ML ($\alpha = 0.0193$), along with a massive 370-fold reduction of the inhomogeneous linewidth broadening. The combined structural, magnetic, and dynamic results indicate that using CoFeB suppresses microstructure-induced magnetic inhomogeneities while staying in the PMA regime. These findings identify CoFeB/Ni MLs as a promising platform for perpendicular spintronic devices, combining low magnetic damping, which is beneficial for coherent magnetization dynamics and long-range magnonic signal propagation, with well-defined domain states.

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I. INTRODUCTION

One of the most prominent next-generation memory technologies is magnetoresistive random-access memory (MRAM), which is considered a key candidate for advancing modern in-memory computing [1,2]. A promising approach to manipulating MRAM cells is the use of spin-transfer torque (STT) [3,4]. A major challenge in STT-MRAM devices is achieving high-density, high-speed operation while maintaining a low critical current density required to switch the free layer. For spin-valve devices with a free layer that has perpendicular magnetic anisotropy (PMA), the switching currents

are smaller than for in-plane devices [5,6]. Well-established PMA systems suitable for use as free and/or reference layers in MRAM cells include Co/heavy-metal magnetic multilayers (MLs) [7–9]. However, a significant drawback of these systems is the spin-pumping effect induced by the strong spin-orbit coupling of the heavy metals. The spin-pumping effect intrinsically increases the magnetic damping, which in turn raises the current density required to manipulate the free layer [10]. One possible way to mitigate this issue is to avoid using heavy metals in the ML. Co/Ni MLs, for example, exhibit relatively low intrinsic magnetic damping [11–13] and high spin polarization [14,15], while maintaining a PMA comparable to that of Co/Pt or Co/Pd MLs [16–19]. While the intrinsic magnetic damping is primarily determined by the constituent materials and governed by spin-orbit coupling [20], contributions from interfaces or neighboring layers are often more pronounced than the intrinsic contribution in real thin films, thereby effectively increasing the measured magnetic damping. To further reduce the magnetic damping, Co can be replaced by an amorphous CoFeB alloy, which

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is well known for its smooth structural growth [21,22], soft magnetic properties [23,24], high spin polarization [25], and very low magnetic damping [26], largely due to a reduced grain boundary formation. Although PMA in CoFe/Ni and CoFeB/Ni MLs has been reported previously [27,28], these studies are largely confined to ultrathin films with total magnetic thicknesses below 10 nm, where interfacial contributions of the seed and cap layer are expected to dominate more. In particular, the PMA reported in thin Pt/[CoFeB/Ni]₃/Pt stacks [28] is likely governed primarily by the interface-induced (or interfacial) anisotropy at the Pt/CoFeB interface [29,30]. Besides these scattered findings, a systematic study of the impact of using CoFeB in the ML stack on its structural, magnetic, and dynamic properties is still missing.

In this study, we address this gap by presenting a comprehensive investigation of the structural and magnetic properties of CoFeB/Ni and Co/Ni MLs, both of which exhibit PMA. The paper is structured as follows. In Sec. II, we describe the material system and experimental techniques in detail. Our results, presented in Sec. III, are organized into six subsections. In Sec. III A, the microstructure and surface morphology are analyzed in real space using scanning transmission electron microscopy (STEM) and atomic force microscopy (AFM). These observations are further validated and complemented by x-ray reflectivity (XRR) and x-ray diffraction (XRD) measurements presented in Secs. III B and III C. In Sec. III D, the static magnetic properties are characterized using a superconducting quantum interference device vibrating sample magnetometer (SQUID-VSM). After subjecting the samples to specific magnetic field histories, the magnetic domain state is examined in Sec. III E via magnetic force microscopy (MFM). Cross-correlation analysis of the resulting domain configurations then quantifies magnetic return-point memory (RPM) of the remanent state and tracks its evolution during field cycling. In Sec. III F, dynamic magnetic properties are probed using ferromagnetic resonance (FMR), allowing precise determination of the uniaxial anisotropy constant K_u , the magnetic Gilbert damping α and the inhomogeneous linewidth broadening. Finally, the key experimental findings are discussed in Sec. IV, and the study is concluded in Sec. V.

II. MATERIALS AND METHODS

Direct current magnetron sputtering is used to fabricate the samples concurrently on thermally and natively oxidized Si(100) wafer pieces with around 100 and 2 nm SiO₂, respectively. The overall structural and magnetic properties are comparable for thermally and natively oxidized Si substrates. A 1.5 nm Ta adhesion layer [31] and a 20 nm Pt seed layer are used as an underlayer for the MLs. In addition to serving as an adhesion layer, the Ta layer also prevents forward sputtering of heavy Pt atoms into the Si/SiO₂ substrate [32], thereby providing a good texture for the Pt seed layer, which is subsequently transferred to the ML [32,33]. The Pt seed layer is used to induce a face-centered cubic (fcc) (111) texture in the MLs, which is beneficial for achieving PMA [34–41]. The samples are capped with a 3 nm Pt layer to prevent oxidation of the ML. The ML stack itself is made up of 50 repetitions of a bilayer consisting of 0.3 nm CoFeB

or Co and 0.6 nm Ni. The high number of repetitions enables better analysis of the texture using structural characterization techniques and leads to the formation of out-of-plane domains at remanence by overcoming the critical thickness for the magnetization reorientation transition [42–44]. Furthermore, possible contributions like spin-pumping or the distinct induced interface anisotropy from the Pt interfaces at the top/bottom of the structure contributes less to the overall magnetic properties. The final sample stack is the following: Si/SiO₂ (2 or 100 nm)/Ta (1.5 nm)/Pt (20 nm)/[FM (0.3 nm)/Ni (0.6 nm)]₄₉/FM (0.3 nm)/Pt (3.0 nm) with FM being either CoFeB or Co.

The samples were grown in an AJA International ATC 2200-V sputter-up system with 2-inch Co, Ni, Pt, and CoFeB (atomic % stoichiometry of 60:20:20) targets. The chamber base pressure is kept below 5×10^{-8} Torr ($< 6.67 \times 10^{-11}$ bar). All depositions are performed at room temperature and 3 mTorr (≈ 4 μ bar) at approximately 20 cm target-substrate-distance with the substrate rotating at 1.25 Hz throughout the deposition.

Bright-field (BF), annular dark-field (ADF), and high-angle annular dark-field (HAADF) STEM imaging was carried out at multiple magnifications using a Talos F200X transmission electron microscope. BF imaging collects the directly transmitted beam and is sensitive to mass-thickness and diffraction contrast, whereas ADF imaging detects scattered electrons and is more sensitive to structural variations. HAADF-STEM, by contrast, emphasizes incoherent high-angle scattering, yielding image intensities that scale approximately with atomic number and enabling direct compositional interpretation.

AFM measurements were performed using a NanoWizard I AFM (JPK Instruments AG) in tapping mode.

All x-ray experiments were conducted with a Rigaku SmartLab thin-film diffractometer. XRD and Rocking curve measurements employed a two-bounce Ge(220) monochromator to isolate the Cu K _{α 1} line and suppress unwanted wavelength contributions. For XRR, a Cu K _{β} filter was used to reduce the influence of the Cu K _{β} radiation. Additionally, in-plane XRD at grazing incidence was performed. The incident angle was fixed at 0.62°, while the detector scanned in the plane parallel to the film surface.

The XRR data were analyzed using GenX 3 [45] (version 3.6.27). The fitting approach used follows a recently developed method for such ML systems and is described in detail in Ref. [32]. The adjusted model used for the samples is described later in the text (see Sec. III B).

Static magnetic measurements in out-of-plane (OOP) and in-plane (IP) geometries are carried out using a Quantum Design MPMS3 SQUID-VSM. A linear background is subtracted from the magnetization curves to achieve a constant saturation. Effective anisotropy values are determined from the area enclosed between the normalized OOP and IP magnetization curves, multiplied by the saturation magnetization [46,47]. The saturation moment is extracted using a geometry-independent moment correction method [48]. The magnetic volume used for magnetization calculations is obtained by multiplying the sample area by the magnetic thickness, which is assumed to equal the ML thickness ([FM (0.3 nm)/Ni (0.6 nm)]₄₉/FM (0.3 nm)).

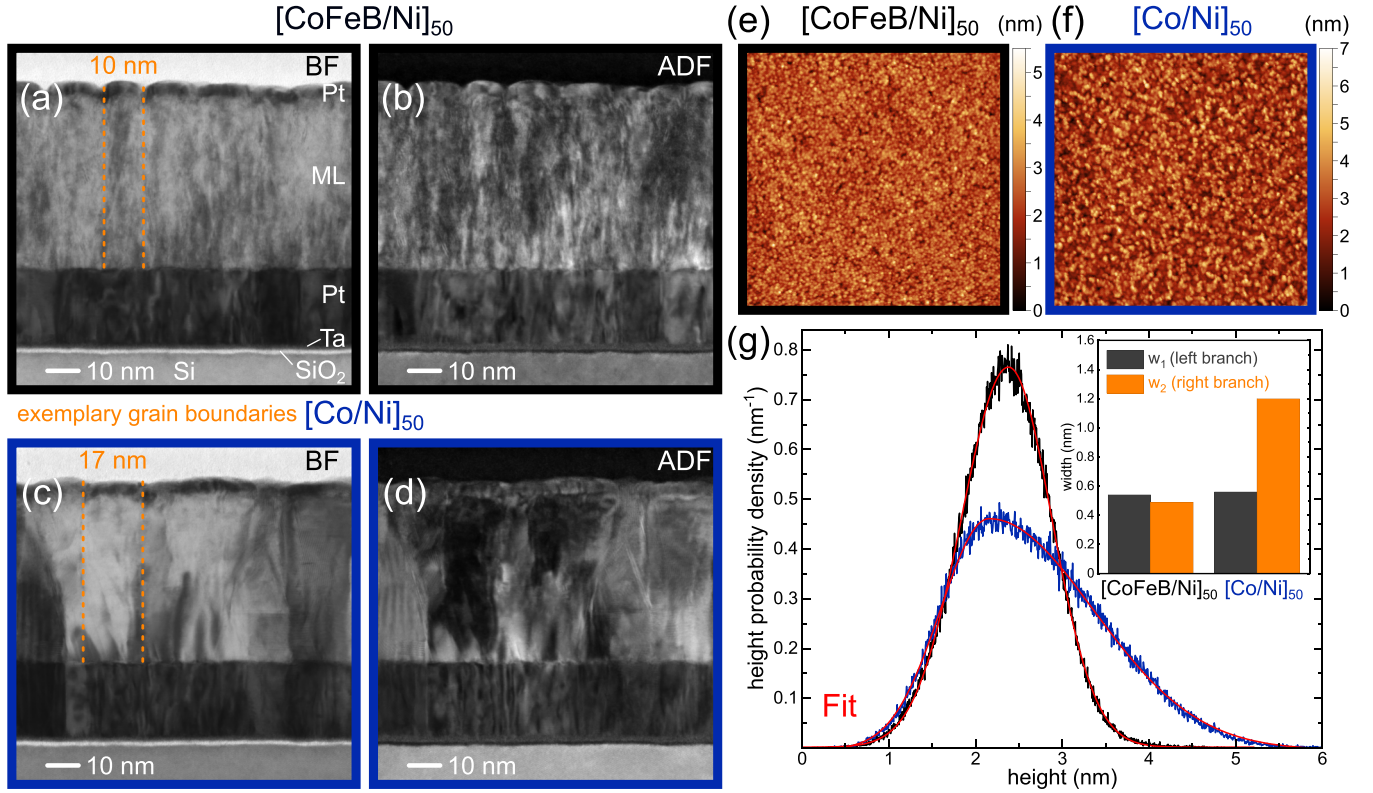


FIG. 1. (a)–(d) Cross-sectional STEM images of the [CoFeB/Ni]₅₀ [black frame, (a) and (b)] and [Co/Ni]₅₀ [blue frame, (c) and (d)] sample using bright-field (BF) and annular dark-field (ADF) detectors. (e) and (f) (1 × 1) μm² AFM images of the samples. (g) Height probability density functions of the AFM images in their assigned colors, with the red curves representing a split-Gaussian fit. The inset shows the widths of the left (w_1) and right (w_2) branches of the fit. The formula for fitting is $y = y_0 + h \cdot \exp\left[-\frac{(x-x_c)^2}{2(w_{1,2})^2}\right]$, with y -axis offset y_0 , peak height h , the center of the Gaussian x_c , and the left and right branch widths $w_{1,2}$ (w_1 for $x < x_c$ and w_2 for $x > x_c$).

Field-free MFM measurements are performed using a Bruker Dimension Icon AFM, with MFM phase images acquired in dual-pass lift mode. Both OOP and IP demagnetization field routines are carried out using an electromagnet integrated into a Microsense EZ7 vector-VSM.

In-field MFM measurements were performed using a NanoScan high-resolution magnetic force microscope (HR-MFM). Microscopy is conducted in noncontact mode, which reduces changes in the sample's magnetization orientation during measurement. The frequency shift is measured, while the phase difference is kept constant for resonant excitation. Before the in-field measurements, the samples were OOP saturated in an external -1.8 T field. In the HR-MFM, an OOP field with the opposite polarity was applied by approaching a permanent magnet installed in the setup. The maximum *in situ* OOP magnetic field is ± 0.45 T. Due to a lateral offset of the magnet with respect to the tip, the magnetic field also exhibits a small IP component at the measurement position.

The dynamic magnetic behavior is investigated using broadband FMR measurements, in which the samples are placed flip-chip on a coplanar waveguide between the magnetic pole shoes and excited magnetically at a fixed microwave frequency provided by a vector network analyzer. The external magnetic field is swept using the electromagnet, and the transmission scattering parameter (S_{21}) is measured at the second port of the vector network analyzer. At FMR,

the magnetic system absorbs microwave energy, which results in a Lorentzian line shape in the transmitted signal. The resonance field and FMR linewidth are extracted from the central maximum and full width at half maximum (FWHM) of the Lorentzian, respectively. To obtain the resonance field and linewidth dependencies, this procedure is repeated over multiple frequencies ranging from 5 to 45 GHz and magnetic fields from 0 to 2.2 T.

Further details on the methods can be found in Sec. I of the Supplemental Material (SM) [49].

III. RESULTS

A. Microstructure and surface morphology

We begin our analysis by examining the microstructure in real space. Cross-sectional STEM images were acquired for both samples on natively oxidized Si substrates. In Figs. 1(a)–1(d), STEM images of the identical sample region are shown using different detectors to highlight variations in the microstructure. Complementary large-scale overview HAADF- and ADF-STEM images are provided in Fig. S1 in the SM [49].

The BF-STEM images of the CoFeB/Ni and Co/Ni sample in Figs. 1(a) and 1(c) reveal columnar grain growth templated by the Pt seed layer. The Pt seed layer also contains occasional voids and diffraction contrast associated with

structural defects, which are also confirmed by the overview HAADF-STEM images in Fig. S1 in the SM [49]. The Ta/Pt underlayer is identical in both samples and forms a smooth, continuous transition from the crystalline Si substrate through the amorphous SiO₂/Ta layers to the crystalline Pt seed layer. Details of the crystallographic structure are discussed in Sec. III C, which provides insight into the average crystallographic structure rather than the local microstructure.

The lateral grain sizes in CoFeB/Ni exhibit a narrower distribution and are shifted towards smaller values, with a mean grain size of approximately 14 nm, compared to 20 nm for Co/Ni (see Fig. S2, SM [49]). This trend is particularly evident in the overview STEM images shown in the SM [49], where Co/Ni exhibits grain sizes ranging from small grains of approximately 6 nm up to larger grains of 42 nm. In contrast, the grain sizes in CoFeB/Ni are distributed within a narrower range of approximately 6–20 nm.

Grain boundaries are visible in both samples (see orange markers in Fig. 1), particularly through the buckling of the Pt cap layer, whose surface morphology follows the underlying grain structure. The weaker grain-to-grain contrast observed in CoFeB/Ni in Fig. 1(a) is likely related to the boron, which modifies the crystallization process by suppressing excessive grain growth and reducing long-range crystallographic ordering (see Sec. III C). Consequently, the reduced grain-to-grain variations and lower degree of texture result in less pronounced diffraction contrast between adjacent grains in the BF-STEM images. Additionally, stacking faults and twinning are observed within individual grains of Co/Ni, whereas such features are not observed in CoFeB/Ni within the investigated regions, as is evident from the ADF-STEM and magnified panel images provided in Fig. S1 in the SM [49].

At the surface, complete capping was achieved, successfully preventing ML oxidation. Nevertheless, oxygen may still diffuse through microstructural defects and locally oxidize the ML film. Such localized oxidation would most likely occur preferentially at grain boundaries, where the formation of a thin passivating oxide layer could self-limit further lateral oxidation into the metallic grains. While the surface buckling is mild and comparable in both samples in Fig. 1, surface differences become apparent in the overview STEM images in the SM [49]. Within the limited field of view, the CoFeB/Ni sample exhibits a smoother surface with more uniform grain heights, whereas the Co/Ni sample shows occasional prominent grains (see, e.g., Fig. S1(c), SM [49]). This observation is further justified by the AFM measurements discussed later.

Focusing on the ML itself, chemical contrast variations in the STEM images in Fig. 1 are barely visible due to the similar mass density of the constituent elements (Co: 8.834 g/cm³, Ni: 8.907 g/cm³ [50], Co₆₀Fe₂₀B₂₀: ≈ 7.347 g/cm³ estimated from the respective mass ratios), but the lattice planes can still be distinguished upon enlargement (see, e.g., magnified panels Fig. S1, SM [49]). The lattice planes within each grain are parallel and oriented in a single direction, with no significant curvature.

Overall contrast variations within or between individual grains occur across different length scales in both samples, with those in Co/Ni occurring on significantly larger ones. ADF-STEM images of the samples are shown in Figs. 1(b)

and 1(d), where contrast mainly arises from scattering associated with structural variations. The same imaging conditions are used for the overview ADF-STEM images in the SM [49]. While Co/Ni exhibits stronger contrast variations primarily at the grain scale (intergranular), in CoFeB/Ni, the scattered intensity varies predominantly on shorter length scales. The latter observation points to slight orientation variations within individual grains (intragranular).

To quantify the surface morphology on a larger scale, $(1 \times 1) \mu\text{m}^2$ AFM scans of the samples on thermally oxidized Si substrates were acquired and are shown in Figs. 1(e) and 1(f), with the corresponding height probability densities in Fig. 1(g). The height probability density is fitted with a split-Gaussian function to account for asymmetries, with the Fig. 1(g) inset showing histograms of the left and right branch widths of the fit to better visualize the grain height distribution.

In agreement with the STEM analysis, both samples exhibit a granular surface structure with distinct morphological differences. In comparison to Co/Ni, the surface grain distribution of the CoFeB/Ni sample appears more uniform, with grains of lower and more narrowly distributed heights, producing a Gaussian height distribution in Fig. 1(g) with a small asymmetry towards lower heights (see inset). For Co/Ni, some more prominently protruding grains above a flatter surface background are visible in both the AFM images and the overview STEM images in the SM [49] as already mentioned, resulting in a non-Gaussian height distribution with a noticeable asymmetry towards larger heights. The extracted root-mean-square (RMS) roughness values σ from the AFM scans are $\sigma_{\text{CoFeB/Ni}} = 0.53$ nm for CoFeB/Ni and $\sigma_{\text{Co/Ni}} = 0.86$ nm for Co/Ni, indicating a surface roughness that is approximately 60% higher for the Co/Ni than for the CoFeB/Ni sample. However, it must be noted that the RMS value may not be the most suitable metric for describing the overall roughness of the Co/Ni sample surface due to the pronounced height asymmetry caused by the prominent higher grains [51]. Grain-size analysis (not shown) reveals lateral surface grain sizes in good agreement with the mean grain sizes obtained from the overview STEM images, confirming the slightly smaller grain size in CoFeB/Ni compared to Co/Ni. Furthermore, the AFM measurements reveal a more irregular surface-height distribution for the Co/Ni sample complementing the real-space analysis.

B. X-ray reflectivity and modeling

To support our real-space structural analysis, we derived average depth profiles of the samples using reciprocal-space XRR, which is sensitive to refractive index variations and, consequently, to electron-density contrasts. Moreover, XRR provides an average of a wider area of the sample than local measurements, such as STEM imaging. Additionally, it was demonstrated that good agreement between the depth profiles of STEM imaging and XRR modeling results can be achieved in a similar Co/Pt ML stack [32].

The XRR profiles for the CoFeB/Ni and Co/Ni sample on thermally oxidized Si substrates are shown in Fig. 2(a). The red lines show the best fits for the XRR data, from which the scattering length density (SLD) depth profiles are extracted

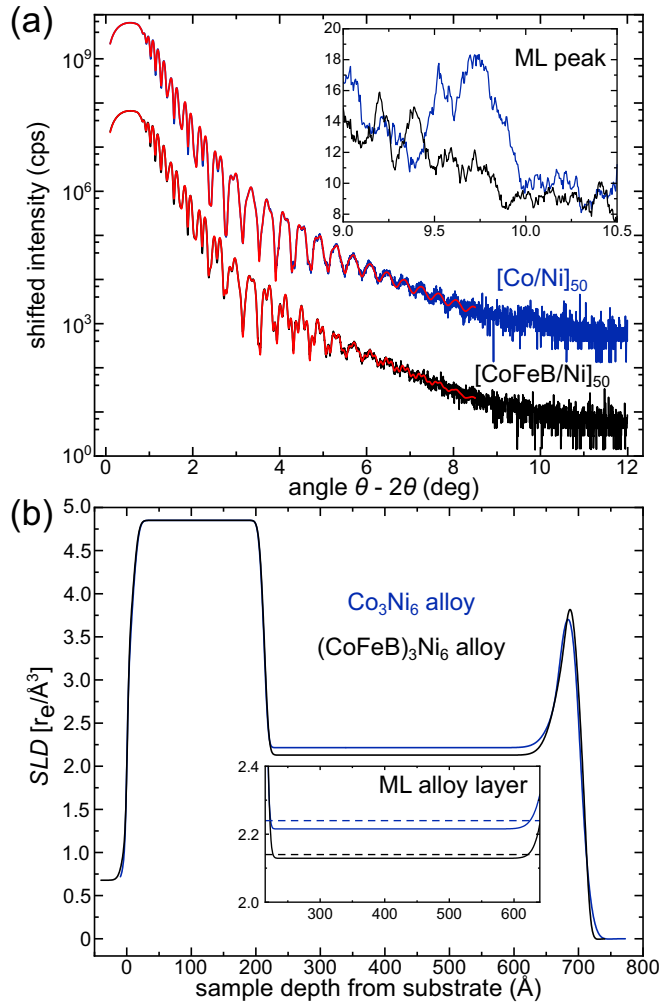


FIG. 2. (a) XRR curves for the $[\text{CoFeB/Ni}]_{50}$ (black) and $[\text{Co/Ni}]_{50}$ (blue) sample on a semilog scale, overlaid with the best fit (red) and vertically shifted for better visualization. The inset highlights the superstructure region (after smoothing) on a linear scale. The (real part) scattering length density (SLD) profiles derived from these fits are shown in (b), where the dashed lines in the inset represent the SLD of a $(\text{CoFeB})_3\text{Ni}_6$ (black dashed) and Co_3Ni_6 (blue dashed) alloy.

[see Fig. 2(b)]. Narrow XRR oscillations correspond to the total film thickness, while wider oscillations are caused by the 20 nm Pt seed layer. Total reflection of the x-rays occurs at around 1° , followed by a strong decrease in intensity according to $1/\sin^4(\theta)$ from the Fresnel equations [52], and further increased by the roughness. The bilayer period d_{ML} of 0.9 nm should result in a superstructure peak at around 9.5° , when using Bragg's law

$$n \cdot \lambda = 2 \cdot d_{\text{ML}} \cdot \sin(2\theta/2), \quad (1)$$

with $n = 1$ and the x-ray wavelength of $\text{Cu K}_{\alpha 1}$ radiation, $\lambda = 1.54 \text{ \AA}$. Because of the low mass density contrast between the constituent elements, as already noted in Sec. III A, only a weak superstructure peak is expected. Due to the noise level in this angular range, a poorly visible superstructure peak can be identified for Co/Ni at the expected position in the full scans and none for CoFeB/Ni. For better visualization,

the XRR curves in the superstructure region are smoothed using adjacent averaging to improve the signal-to-noise ratio as shown in the Fig. 2(a) inset. From the best fits, we determine the total thickness of the ML to $(45.6 \pm 0.2) \text{ nm}$ for Co/Ni and $(45.8 \pm 0.2) \text{ nm}$ for CoFeB/Ni, which agree well with the nominal thickness of approximately 45 nm. For the Pt seed layer, we determine an average thickness of $(20.3 \pm 0.2) \text{ nm}$, which also shows excellent agreement with the nominal value.

To extract the SLD profile from the XRR measurements, modeling and refinement are required. An effective approach to model such thick MLs is to approximate the ML structure as an alloy while excluding the superstructure peak, as recently introduced by Ehrler *et al.* [32]. Furthermore, the inclusion of alloylike transition layers (TLs) at the interfaces with the substrate, seed, and cap layer helps to represent the actual interlayer transitions, acting as an effective medium approximation. This approach is further motivated by the observation that sputtered ML structures can behave more like an effective alloy than an ideal, perfectly layered system [53]. The ML alloy mass compositions are estimated using the respective thickness ratios of the individual layers. We approximate the 50 bilayer repetitions of Co/Ni with a 3:6 Co_3Ni_6 alloy, proportional to the bilayer thickness ratio. For CoFeB/Ni, taking the CoFeB target composition of 60:20:20 into account, the resulting alloy composition reads $(\text{CoFeB})_3\text{Ni}_6$, which expands to $\text{Co}_3\text{FeB}_3\text{Ni}_{10}$. Using the literature values for the components, we estimate the “bulk” density of those alloys. To suppress any residual contributions arising from the ML periodicity, the fitting range is restricted to angles below 8.5° . We therefore model our samples as follows:

substrate(SiO_2) / substrate-underlayer TL / Ta / Pt_{seed} / seed-ML TL / alloy / ML-cap TL / Pt_{cap} / cap-air TL.

SiO_2 can be set as substrate in good approximation due to its thickness of 100 nm. For the substrate-underlayer TL, a Ta_2O_5 layer with a mass density of 8.2 g/cm^3 is used to account for the possible formation of TaO_x at the interface in sputtered films [54] and to achieve an optimal fit of the substrate-underlayer transition [32]. For each layer, the roughness and the thickness are refined. We fixed the densities of SiO_2 , Ta_2O_5 , Ta, and Pt_{seed} to their bulk values, fitting all the others.

The modeled SLD profiles are presented in Fig. 2(b), with an inset highlighting the ML alloy layer. The dashed lines indicate the calculated SLD values of “bulk” $(\text{CoFeB})_3\text{Ni}_6$ and Co_3Ni_6 alloys. Fitted and calculated alloy densities are in good agreement, supporting the validity of the approximation. Both SLD profiles are nearly identical, visually confirming that the two samples have the same thickness. The only notable difference occurs at the cap/air transition, where the Co/Ni sample shows a slightly lower Pt cap density and a broader right shoulder, indicative of increased surface roughness, which is in agreement with the STEM and AFM measurements in Sec. III A. The cumulative roughness of the entire layer stack is reflected best in the fitted SLD profile at the sample surface. To circumvent the ambiguity of the roughness fitting parameters due to the surface TL, we fit the surface transition directly of the SLDs using an error function to determine the roughness from the error function's

width [41]. The uncertainties are estimated from the fitting error and from variations in the fitting range to account for the complex transition shape. We obtain fitted widths of $\sigma_{\text{CoFeB/Ni}} = (0.7 \pm 0.1) \text{ nm}$ and $\sigma_{\text{Co/Ni}} = (1.2 \pm 0.2) \text{ nm}$ for CoFeB/Ni and Co/Ni, respectively. This is a similar relative increase to that of the RMS roughness derived from the AFM measurements. The absolute magnitude of the roughness differs, which is known from the literature and can have different reasons, such as the large difference in investigated area and the non-Gaussian height distribution [55]. As already pointed out in Sec. III A, the RMS value may not be the most representative measure of the surface roughness, as it is influenced by the asymmetric height distribution.

C. Crystallographic characterization

For further structural characterization, specular OOP XRD and Rocking scans of the samples on natively oxidized Si substrates are presented in Fig. 3. Two prominent Bragg reflections are visible for both samples in the displayed angle range: (i) the Pt(111) reflection from the 20 nm seed and 3 nm cap layer at approximately 39.6° and (ii) the ML reflection at around 44.5° . An additional weak shoulder is seen at approximately 34.7° , which can be attributed to a satellite peak (ML-1) arising from the periodicity of the ML. The satellite features are only weakly developed in both samples, similar to the superstructure peaks in XRR (see Fig. 2(a), but only for Co/Ni). The satellites appear at approximately 10° lower angles relative to the ML(111) peak, corresponding to a ML period of approximately 0.9 nm, as determined from Eq. (1) for the first order ($n = 1$), where $2\theta \approx 10^\circ$. This value agrees well with the ML period extracted from the XRR analysis.

Returning to the main Bragg reflections, both samples exhibit a pronounced fcc(111) texture for the Pt and ML. Other lattice directions are not observable in the OOP XRD scans. For both Pt seed layers, the Bragg reflections are close to their respective bulk values, as indicated by the dashed-dotted line in Fig. 3, with the shifts explainable by a small deposition-induced stress. Their XRD intensities, as well as the Rocking scan intensities shown in the top left inset of Fig. 3, and their corresponding FWHM values extracted from Voigt fits, as expected, comparable for both samples. This aligns well with our STEM and XRR analyses, which suggest that the thickness and surface quality of the underlayer are similar in both samples. This also highlights the good reproducibility of the sputtering process used to fabricate the samples.

The differences between the two samples become more apparent in the crystal structure of the ML. For Co/Ni, its peak position agrees well with the expected “bulk” value, which can be estimated by a linear interpolation of the lattice plane distances of fcc Co(111) (2.046 Å, ICSD-622435) and fcc Ni(111) (2.035 Å, ICSD-646088). This yields an effective ML(111) lattice parameter of 2.039 Å, corresponding to a Bragg angle of $2\theta = 44.37^\circ$ using Eq. (1) with $n = 1$, which is also displayed as a dash-dotted line in Fig. 3. For CoFeB/Ni, the boron in the sample should prevent good crystallization. Due to the Pt seed templating a fcc(111) texture and the boron not being directly deposited into the Ni, we still observe a ML reflection for the CoFeB/Ni ML. The ML(111) peak

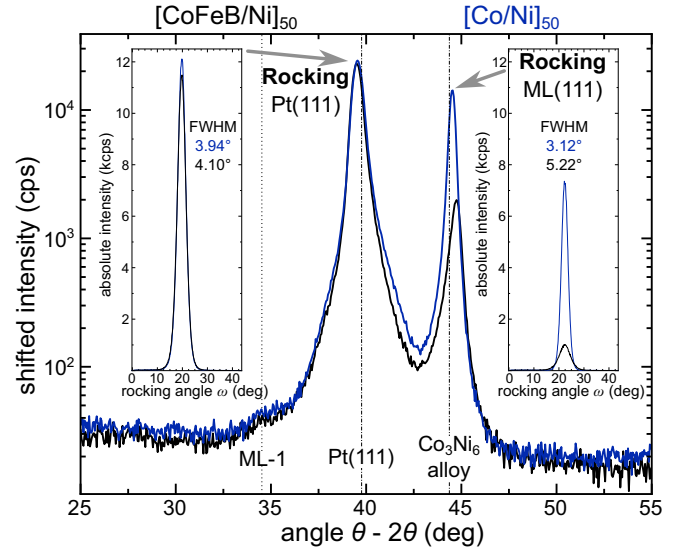


FIG. 3. OOP XRD scans for the $[\text{CoFeB/Ni}]_{50}$ (black) and $[\text{Co/Ni}]_{50}$ (blue) sample. The insets show Rocking scans of the Pt(111) and ML(111) Bragg reflections on a linear scale, with the corresponding FWHM extracted from Voigt fits. Dash-dotted lines indicate the bulk reference reflections, which are labeled at the bottom of the XRD scans and described in the text.

of CoFeB/Ni is shifted to slightly higher angles compared to Co/Ni, which can be attributed to lattice compression induced by the addition of Fe. Moreover, the ML diffraction and Rocking scan (top right inset in Fig. 3) intensities differ drastically between the two samples. Using CoFeB in the ML leads to a one-order-of-magnitude reduction in intensity and a significant broadening of the diffraction and Rocking scan profile.

To further investigate the crystal structure, we also performed IP XRD at grazing incidence, which is sensitive to lattice planes parallel to the film surface. In this geometry, the fcc $\{2\bar{2}0\}$ planes should be visible, since they are perpendicular to the OOP fcc(111) planes. For the IP XRD scans in Fig. S3 in the SM [49], clear fcc $\{2\bar{2}0\}$ reflexes are visible for the Pt and ML. For the Co/Ni sample, additional weak fcc $\{1\bar{1}1\}$ contributions are apparent in the IP XRD scans, corresponding to a minor fcc(220) OOP phase, hidden in the OOP scan by the large intensity from the Si substrate (not shown).

By fitting the XRD scans using a symmetric Pearson VII function and analyzing their FWHM, the coherence length was estimated using the Scherrer equation (shape factor $K = 0.94$). The resulting OOP coherence lengths are approximately 29 nm for the Co/Ni and 20 nm for the CoFeB/Ni ML, both of which are smaller than the total layer thickness ($\approx 45 \text{ nm}$) [56]. The IP XRD measurements show also a decrease in coherence length from roughly 12 to 4 nm, going from Co/Ni to CoFeB/Ni. Note that OOP XRD probes the vertical crystallite size along the film normal, whereas IP XRD probes the lateral grain size. The larger values obtained from the OOP XRD further demonstrate that the grains are elongated along the growth direction, resulting in columnar growth. In contrast, the shorter IP values for the CoFeB/Ni ML reflect the reduced mean lateral grain size from the STEM

analysis. Collectively, we find good agreement in the relative proportions between the IP coherence lengths and the lateral grain sizes identified in the STEM images, even though the absolute values do not match exactly, since the crystalline coherence length is only a lower estimate for the grain size. The Scherrer method also does not account for growth-induced strain and instrumental broadening, which leads to further deviations. In Co/Ni, the structural compatibility between Co and Ni promotes coherent growth across interfaces, resulting in an extended OOP coherence length. In contrast, the use of CoFeB disrupts crystallographic order and introduces local structural disorder, reducing the coherence length. At the same time, CoFeB suppresses grain coarsening and enhances nucleation density, leading to laterally smaller grains. For the Pt seed layer, an IP coherence length of 9 nm for both samples is determined. This indicates that grain coarsening occurs in Co/Ni (12 nm), whereas in CoFeB/Ni (4 nm) new grain boundaries are everywhere nucleated. We note that the local structural differences observed in the STEM images between Co/Ni and CoFeB/Ni (e.g., grain size and stacking faults) do not contradict the higher degree of crystallinity for the Co/Ni sample indicated by XRD. The structural variations at the nanoscale are averaged out in XRD measurements and therefore do not necessarily affect the overall crystallographic quality. While boron suppresses grain coarsening in CoFeB/Ni, it also reduces the degree of crystallographic texture.

Having established the structural results from real- and reciprocal-space analyses, we now examine their impact on the magnetic properties.

D. Static magnetic characterization

The SQUID-VSM magnetic hysteresis curves for the OOP (light blue) and IP (yellow) geometries, as well as some exemplary MFM phase images at remanence after saturation and demagnetization, are shown in Figs. 4(a)–4(d), respectively. The OOP demagnetization field routine produces an isotropic labyrinthlike stripe domain pattern. In contrast, IP demagnetization favors the formation of partially ordered stripe domains aligned with the applied field. The demagnetized states exhibit approximately equal populations of up- and down-magnetized domains. The magnetic parameters obtained from the SQUID-VSM measurements are depicted in Table I.

The saturation magnetization of $M_s = 836$ kA/m for the Co/Ni sample agrees well with the expected value of $M_s(\text{Co}_3\text{Ni}_6) = 805$ kA/m, which is calculated from the weighted average of the bulk M_s values of Co (1440 kA/m) and Ni (488 kA/m) [57]. Because boron is included, estimating the expected saturation magnetization for the CoFeB/Ni sample is not straightforward. A weighted average of the bulk magnetizations of Co, Fe (1710 kA/m [57]), and Ni, assuming no magnetic or volume contribution from boron, yields a stoichiometry of $\text{Co}_3\text{Fe}_1\text{Ni}_{10}$ with a magnetization of 727 kA/m. On the other hand, it is known that the average magnetic moment of the CoFeB alloy is reduced through a boron dilution effect [58,59]. As a simple estimate, we assume that boron suppresses the moment of an equal amount of Co and Fe, proportionally to their ratio, while contributing no magnetic moment or volume itself. The effective magnetic

TABLE I. Magnetic parameters obtained from SQUID-VSM measurements: saturation magnetization M_s , effective anisotropy energy density K_{eff} , and OOP-normalized remanent magnetization M_r/M_s . The bracketed number expresses the uncertainty in the final digits.

sample	M_s (kA/m)	K_{eff} (kJ/m ³)	M_r/M_s (%)
CoFeB/Ni	673(70)	−35(4)	0.03
Co/Ni	836(74)	134(12)	6.97

stoichiometry becomes $\text{Co}_{2.25}\text{Fe}_{0.75}\text{Ni}_{10}$, yielding a saturation magnetization of 672 kA/m. We see that the latter approach matches the experimental value more closely, although both expected values are within the experimental margin of error.

For both samples, the OOP loops show stripe-domain-mediated magnetization reversal [60,61]. This process is characterized by domain nucleation (steep onset), a linear regime associated with domain breathing, and domain annihilation (steep termination). The corresponding changes in magnetization are also clearly reflected in the first derivative of the hysteresis loop, as shown in Fig. S4(b) in the SM [49]. A larger enclosed hysteresis area in Co/Ni results from the lower nucleation field and higher saturation field than in CoFeB/Ni. The CoFeB/Ni loop nearly closes at around ± 0.1 T, leading to an almost zero OOP-normalized remanent magnetization of $M_r/M_s = 0.03\%$. Thus, strikingly, already after OOP saturation, a demagnetized remanent state is achieved (see Fig. 4(c), first column). The reduced saturation magnetization M_s for CoFeB/Ni (see Table I) lowers the demagnetization energy, which would generally favor domain configurations that are more readily stabilized by pinning centers [62]. However, this behavior is not seen here. A similar reduction in domain-wall pinning caused by boron incorporation has previously been reported in CoB-based OOP systems [47,63,64]. Additionally, the ring-shaped FFT, which is shown in the lower right corner of the imaged domain state, confirms this well-defined, isotropic labyrinth stripe domain pattern for CoFeB/Ni. After OOP demagnetization, a similar labyrinthlike stripe domain state with additional bubble domains of both polarities is noticeable. After OOP saturation and OOP demagnetization, the domain periods, extracted from the radius of the ring-shaped FFT, are identical within the experimental uncertainty at 147 and 153 nm, respectively.

The Co/Ni sample shows an OOP-normalized remanent magnetization of $M_r/M_s = 6.97\%$, and is in a more disordered and irregular stripe-domain state at remanence after OOP saturation (Fig. 4(d), first column). Only the OOP demagnetization routine transforms this state into an isotropic, labyrinthlike stripe domain state, albeit with a much broader domain-width distribution than in the CoFeB/Ni sample (broader FFT ring width). Following OOP saturation and demagnetization, the distinct average domain periodicities are 235 and 192 nm, respectively.

The IP loops exhibit S-shaped behavior, with some hysteresis arising from the switching of Bloch-type domain-wall centers [65]. The overall slopes (between the saturation fields) differ for the two samples, indicating different magnetic anisotropy energies. For CoFeB/Ni, the slope is steeper, while for Co/Ni, it is flatter, indicating much more

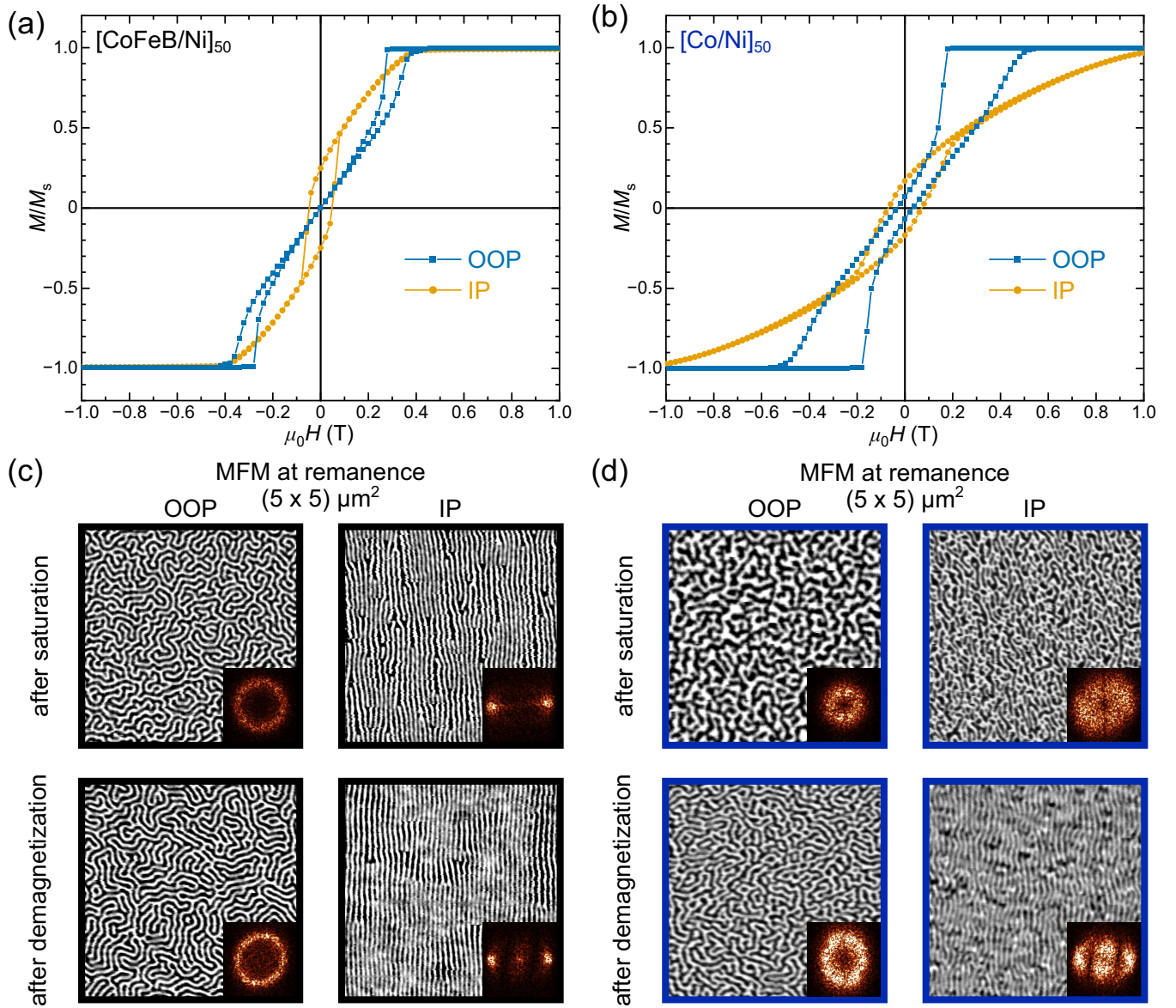


FIG. 4. (a) and (b): SQUID-VSM hysteresis curves for out-of-plane (OOP, light blue) and in-plane (IP, yellow) geometries from the [CoFeB/Ni]₅₀ and [Co/Ni]₅₀ sample. (c) and (d): $(5 \times 5) \mu\text{m}^2$ MFM phase images of the [CoFeB/Ni]₅₀ (black frame) and [Co/Ni]₅₀ (blue frame) sample at remanence. The first rows are after saturation, and the second rows are after demagnetization, with the respective left and right columns having the field applied OOP and IP. The contrast of the MFM phase images is adjusted for each image to improve visibility. The insets show the $(22 \times 22) \mu\text{m}^{-2}$ Fast Fourier Transforms (FFTs) of the MFM phase images, each with the same contrast scale.

pronounced hard-axis behavior than for CoFeB/Ni. Shortly after remanence, the hysteresis curve of CoFeB/Ni exhibits a sudden change in magnetization, suggesting a more unrestricted switching of Bloch-type domain-wall centers. In contrast, for Co/Ni, the transition is more gradual. The Co/Ni sample is not saturated within the displayed field range, and therefore a wider field range, along with the corresponding first derivatives of the IP loops for both samples, is shown in Fig. S5 in the SM [49]. Additionally, for Co/Ni, two distinct peaks are observed in the IP derivative [65], whereas for CoFeB/Ni the abrupt switching of the domain-wall centers manifests as a single peak.

Due to their distinct response to IP fields, we could expect differences in the domain states after IP field treatment. Notably, the MFM images acquired at remanence following

IP saturation and IP demagnetization (Fig. 4(c), second column) for the CoFeB/Ni sample both exhibit a well-aligned stripe domain state, leading to comparable FFT patterns. Upon IP demagnetization, the domain periodicity at remanence is 118 nm, whereas it is longer at 128 nm following saturation. For Co/Ni (Fig. 4(d), second column), only a weakly aligned stripe domain pattern is visible after IP demagnetization with a period of 135 nm. Additionally, another frequency component orthogonal to the stripes appears in the FFT, originating from the recurring forking of the domains. After IP saturation, a mixed phase of elongated stripes and bubble domains is visible. The ratio of the domain periods of the IP and OOP demagnetized state is slightly lower for Co/Ni ($135 \text{ nm}/192 \text{ nm} = 0.70$) compared to CoFeB/Ni ($118 \text{ nm}/153 \text{ nm} = 0.77$). For Co/Pt MLs exhibiting PMA

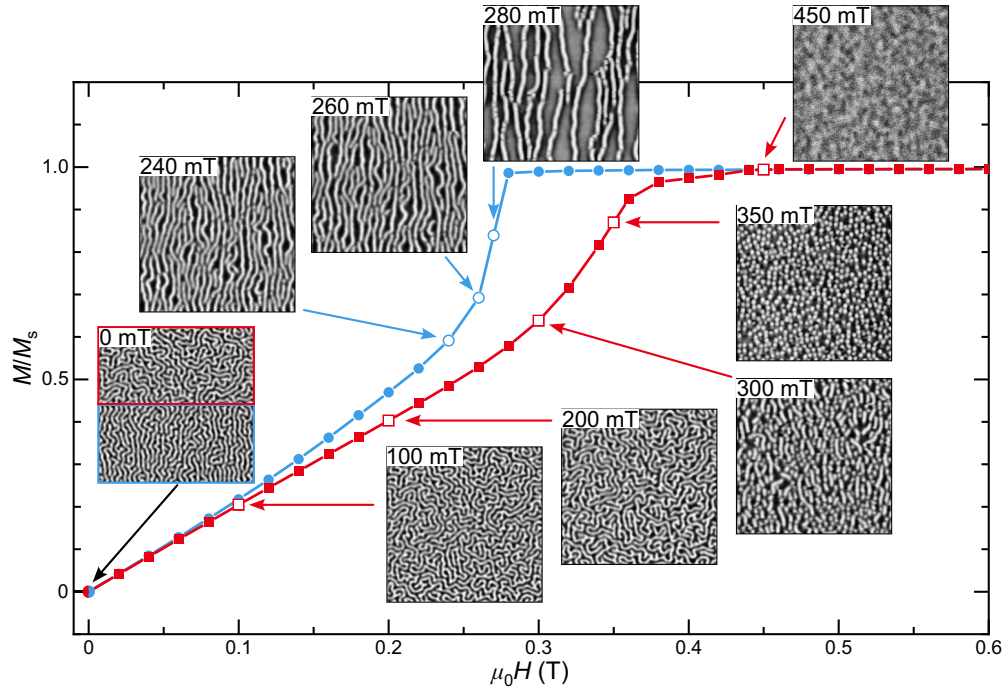


FIG. 5. $(5 \times 5) \mu\text{m}^2$ MFM frequency shift images recorded along the major OOP hysteresis loop [from Fig. 4(a)] of the $[\text{CoFeB}/\text{Ni}]_{50}$ sample. The ascending branch of the loop is indicated in red and the descending branch in light blue. The magnetic states at different applied fields (open symbols) are probed with in-field MFM, starting from remanence after external OOP saturation (0 mT, red frame), proceeding to positive saturation (at 450 mT), and finally returning to remanence (0 mT, light blue frame). A small, parasitic IP magnetic field from the magnet setup, oriented vertically in the MFM image, biases the domains to align along this direction.

and domain patterns comparable to those of the Co/Ni sample, a similar domain-period ratio of approximately 0.7 has been reported [66].

An estimate of the effective anisotropy energy density K_{eff} is given in Table I. However, the area method provides only a rough estimate because of the complication in determining the anhysteretic $M(H)$ curve [47]. A more direct determination of the uniaxial anisotropy constant K_u is obtained from FMR measurements, as the samples are magnetically saturated during the measurement, thereby eliminating contributions from domains (see Sec. III F). The resulting values of $K_{\text{eff}} > 0$ for Co/Ni and $K_{\text{eff}} < 0$ for CoFeB/Ni imply that the uniaxial anisotropy overcomes the shape anisotropy in the former, whereas the latter remains dominated by the shape anisotropy. Despite the negative K_{eff} for CoFeB/Ni, indicating that a uniformly magnetized state favors in-plane orientation, remanent OOP domains are observed in Fig. 4(c). This is not contradictory, as in the weak-PMA regime the reduction in magnetostatic energy associated with, for example, stripe or labyrinthlike stripe domain formation can outweigh the domain-wall energy, stabilizing OOP multidomain states despite $K_{\text{eff}} < 0$ [42,43,67–69].

E. Mapping magnetic reversal using MFM

To determine the magnetic-field-dependent domain states in CoFeB/Ni, in-field MFM is used to image the magnetic configuration at selected external magnetic field values. In Fig. 5 the first quadrant of the major OOP hysteresis loop of the CoFeB/Ni sample [see Fig. 4(a)] is shown, together with the corresponding $(5 \times 5) \mu\text{m}^2$ MFM frequency shift images

that probe the domain configurations along the ascending (red) and descending (light blue) branch of the loop. The domain configurations at remanence after previous external OOP saturation (red frame) and after returning from the descending hysteresis branch (light blue frame) are highlighted by their respective colors. On the hysteresis loop, open symbols indicate the magnetic field values at which the domain states are probed along the corresponding branches, and the arrows mark the associated MFM images. We note that magnetic field values are subject to uncertainties arising from magnet calibration, which increase with field strength and are estimated to be on the order of a few mT at low fields and several tens of mT at higher fields.

Before placing the sample into the HR-MFM, it is exposed to an OOP field above saturation in an electromagnet to achieve microscopic saturation. Then, the first MFM image within the red frame (at remanence) in Fig. 5 is recorded. The measured state coincides with that seen in Fig. 4(c) (first column), where an isotropic labyrinth stripe domain state is formed after OOP saturation. At 100 mT, we are still in the linear regime of the domain breathing mode. In this regime, the domains broaden or narrow depending on their orientation relative to the external field. Subsequently, at 200 mT, the labyrinth stripe domain pattern remains recognizable, but it becomes less dense while the dark domains expand. The annihilation of the labyrinth stripe domain state sets in at around 300 mT, where the pattern breaks up into isolated stripe and bubble domains. Upon further increasing the field to 350 mT, a fully developed and dense bubble-domain state is apparent. The weak yet still present tip-sample interaction

removes some bubble domains, as evidenced by the dark areas in the image at 350 mT. The average bubble domain diameter is approximately 114 nm, with a density of 2960 domains per $100\ \mu\text{m}^2$ (see Fig. S6(a), SM [49]). Additionally, this pure bubble domain state manifests as an additional hump in the first derivative of the SQUID-VSM OOP reversal loops, as shown in Fig. S4 in the SM [49]. To determine the stability of this bubble domain state without an external field, a remanent MFM image is taken after the application of 350 mT, and shown in Fig. S6(b) in the SM [49]. The nearly complete and dense bubble domain state is not fully preserved at remanence, as some bubble domains reconnect and transform into elongated stripe domains, likely due to the small, parasitic IP field of the HR-MFM magnet. Yet, the overall pattern remains dominated by isolated bubble domains. After cycling between 0 and 350 mT, this bubble and remanent bubble-dominated domain state can also be almost fully replicated.

It is noted that the local MFM image at 450 mT shows only a weak granular signal and no residual domains. However, some regions of the sample may remain unsaturated, as microscopic saturation is reached at fields higher than the saturation field indicated in the hysteresis curve [61].

On the descending branch, at approximately 280 mT (below the nucleation field), the rapidly developed stripe domains can be seen. Typically, in MLs with PMA, isolated bubble domains first nucleate at structural defects and subsequently burst into stripe-domain clusters [60,61]. Note that the domains are vertically aligned in the MFM image due to the setup's parasitic IP magnetic field component. The MFM tip field moves and removes certain stripe domains during imaging, as the bright domains correspond to repulsive tip-sample interaction. As the external magnetic field is further reduced, the stripe domains nucleate more densely, and the domain period decreases as more domains emerge in empty areas. Eventually, the system develops into a stripe-domain state at remanence, where the previously quite strong vertical alignment of the stripe domains is somewhat reduced due to the close packing of domains.

For the Co/Ni sample, we could measure only the ascending branch (not shown) because full saturation was not achievable with our setup. On this branch, a noticeable breathing of the stripe domains is observed, which transformed into a bubble domain state shortly before saturation. This is similar to CoFeB/Ni and consistent with previously reported stripe domain field reversals [60,61,65,67].

Furthermore, we probe the return-point memory (RPM) of our samples using in-field MFM by tracking how domain states recover when the magnetic field returns to previously visited values. While RPM is often studied using synchrotron-based methods, which provide high spatial resolution and element-specific contrast [70,71], in-field MFM offers a practical alternative. To analyze how individual domain states are preserved in our samples, we acquired MFM images at remanence following different minor-loop field cycles. First, the measurement starts at remanence after external OOP saturation. Subsequently, progressively higher external fields are applied along the ascending branch of the OOP hysteresis curve, followed by a return to remanence, at which point an MFM image is taken. The microscopic domain memory is quantified using a generalized standard cross-correlation (Pearson correlation coefficient ρ) between two random vari-

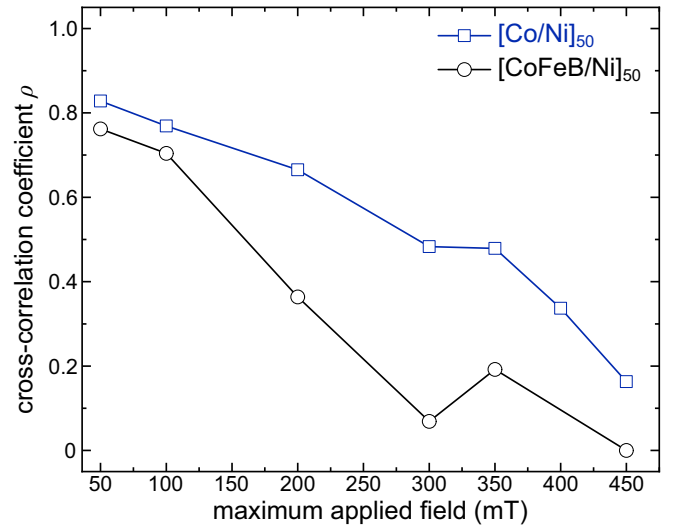


FIG. 6. Evolution of the cross-correlation coefficient ρ under the maximum applied field for the [CoFeB/Ni]₅₀ (black circles) and [Co/Ni]₅₀ (blue squares) sample. After applying an external magnetic field, an MFM image is acquired at remanence and correlated with the previously measured domain state. After each data point is acquired, the next minor loop field is applied directly.

ables a and b [70], defined as

$$\rho(a, b) = \frac{\text{Cov}(a, b)}{\sqrt{\text{Var}(a)\text{Var}(b)}}, \quad (2)$$

with Cov being the Covariance, Var the Variance, and a and b refer to our MFM images. Therefore, if the domain patterns are identical, then the value of ρ is 1. If the patterns are completely uncorrelated, then the value of ρ is 0. The cross-correlation coefficient ρ between consecutive remanent MFM images is a statistical quantity that depends on the finite number of pixels in each image. The error of the cross-correlation coefficient ρ is estimated using its standard error. For our data, the standard error is $\Delta\rho \lesssim 0.03$, which is small compared to the observed variations in ρ and is therefore neglected. MFM images are analyzed using a custom Python script (version 3.14.3) [72], where the images are first converted to gray scale and then binarized by assigning each pixel a value of either +1 or -1, depending on whether its intensity is above or below the mean. Binarizing the MFM images suppresses contrast variations and measurement noise, enabling a more robust comparison of the magnetic domain configurations. The binarized MFM images are shown in Fig. S7 in the SM [49].

In Fig. 6, for both samples, the evolution of ρ in dependence on the maximum applied field is shown. Within the domain breathing regime, up to 100 mT for CoFeB/Ni, the remanent state is reproducible, as indicated by the high ρ values of 0.76 and 0.70 for maximum external fields of 50 and 100 mT, respectively. This behavior is further reflected in the linear slope of the ascending hysteresis branch within this field range in Fig. 5. A significant drop in ρ occurs when the maximum applied field is increased further, suggesting a loss of microscopic RPM and reversibility, although the qualitative appearance remains similar at 200 mT (see Fig. S7, SM [49]). As the external field approaches saturation, the cross-correlation coefficients trend towards zero. Remarkably,

when the pure bubble domain state at 350 mT (see Fig. 5) is reached, ρ rises marginally to 0.19 again. This indicates that some bubble domains remain microscopically unchanged, as also demonstrated in Fig. S6(b) in the SM [49]. While the bubble domain state is not completely preserved upon removal of the external field, it may be stabilized at remanence through appropriate magnetic-field cycling [67,68].

For the Co/Ni sample, domain states are more strongly preserved during minor-loop field cycling. There, ρ shows consistently higher values than for CoFeB/Ni, indicating stronger RPM. Still, the overall trend shows that the RPM is progressively lost as the system approaches saturation. An almost pure bubble domain state is also observed in Co/Ni at 400 mT (not shown). However, it does not persist when the field returns to remanence. This contrasts with CoFeB/Ni, where an increasing correlation coefficient is determined at 350 mT by retaining some bubble domains at remanence. This behavior is most likely caused by differences in PMA strength between the two samples. CoFeB/Ni exhibits a negative K_{eff} , indicating a relatively low domain-wall energy, which might prevent the merging of some bubble domains at zero field. In contrast, Co/Ni has a positive K_{eff} value, leading to a reconfiguration of the domain structure at remanence due to the higher domain-wall energy. Recent studies on Co/Pt MLs have demonstrated the stabilization of magnetic bubble domain states both in systems dominated by the shape anisotropy [67,68] and in systems with strong PMA [73]. However, in the latter case, the bubble domain state cannot be readily stabilized without tilting the external magnetic field away from the film normal, which is consistent with our RPM observations in the Co/Ni sample.

In addition to the minor loops, we performed a second field protocol for the CoFeB/Ni sample in the HR-MFM setup (for Co/Ni, saturation could not be reached). We cycled between remanence $\rightarrow$ OOP saturation $\rightarrow$ remanence and compared the remanent state within each field cycle. The cross-correlation coefficient in dependence of the field cycle is shown in Fig. S8 in the SM [49]. An almost zero ρ indicates low microscopic domain memory in the CoFeB/Ni sample for this specific field routine, suggesting that the domain pattern undergoes substantial reconfiguration during each cycle.

F. Magnetization dynamics

To characterize the magnetization dynamics of the samples, FMR measurements are performed. The magnetic field is swept in the OOP direction at a constant exciting microwave frequency. Fitting the signals by complex Lorentzian curves yields the resonance field ($\mu_0 H_{\text{res}}$) and linewidth ($\mu_0 \Delta H_{\text{pp}}$). These parameters are determined for all the intermediate frequencies and are plotted in Figs. 7(a) and 7(b), respectively. For the OOP geometry, $\mu_0 H_{\text{res}}$ follows the expression known as the Kittel equation for thin films [74–76]:

$$f_{\text{res}} = \frac{\gamma}{2\pi} (\mu_0 H_{\text{res}} - \mu_0 M_{\text{eff}}), \quad (3)$$

where

$$\mu_0 M_{\text{eff}} = \mu_0 M_s - \frac{2K_u}{M_s}. \quad (4)$$

Here, $\gamma = g\mu_B/\hbar$ is the gyromagnetic ratio, with the Landé g factor, μ_B the Bohr magneton, $\hbar$ the reduced Planck con-

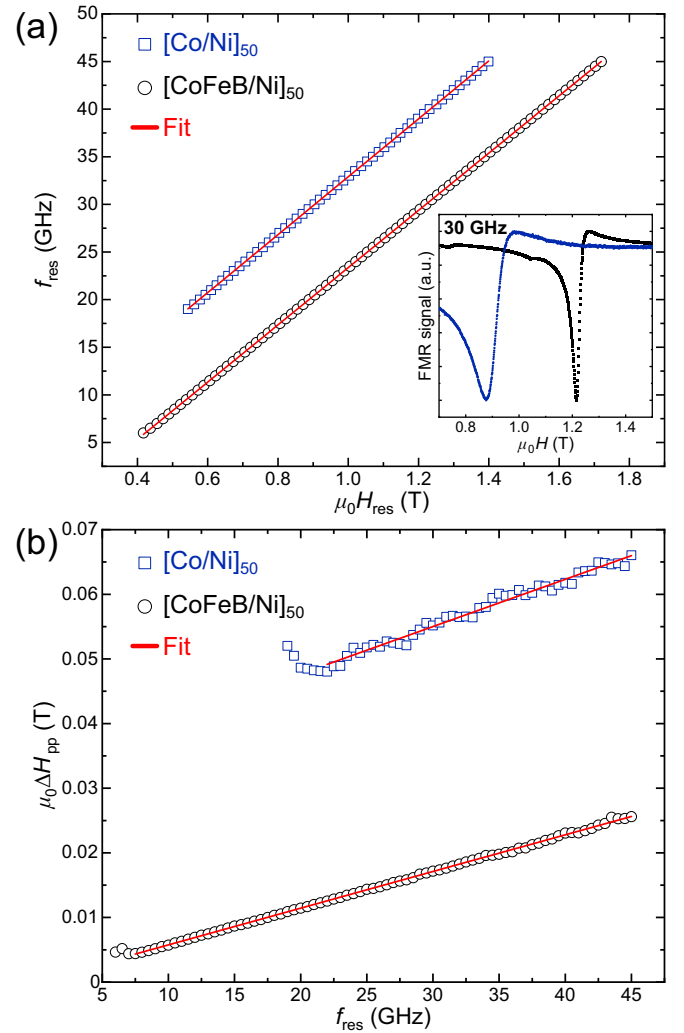


FIG. 7. (a) Resonance frequency f_{res} dependence on the resonance field $\mu_0 H_{\text{res}}$ for the [CoFeB/Ni]₅₀ (black circles) and [Co/Ni]₅₀ (blue squares) sample. The inset shows exemplary raw FMR signals at $f = 30$ GHz. (b) Corresponding linewidth $\mu_0 \Delta H_{\text{pp}}$ dependence on f_{res} for both samples. The red fits follow Eqs. (3) and (5) for (a) and (b), respectively.

stant, and the effective magnetization $\mu_0 M_{\text{eff}}$, which ideally contains all anisotropy contributions. Fitting the experimental $f_{\text{res}}(H_{\text{res}})$ dependency shown in Fig. 7(a) using Eq. (3) allows the effective magnetization $\mu_0 M_{\text{eff}}$ and the Landé g factor to be determined from the intercept and slope of the linear relation, respectively. Using Eq. (4), the uniaxial anisotropy constant K_u can be calculated from the M_s and $\mu_0 M_{\text{eff}}$ values given in Tables I and II.

The Gilbert damping parameter α is extracted from the frequency dependence of the FMR linewidth ΔH_{pp} [see Fig. 7(b)] as

$$\Delta H_{\text{pp}} = \Delta H_{\text{pp}}^0 + \frac{2}{\sqrt{3}} \frac{2\pi\alpha}{\gamma} f_{\text{res}}, \quad (5)$$

where ΔH_{pp}^0 represents the frequency-independent inhomogeneous broadening. The parameters obtained from these relations are displayed in Table II.

In Fig. 7(a), the resonant frequency f_{res} is plotted over the resonant field $\mu_0 H_{\text{res}}$ for the samples. Furthermore,

TABLE II. Magnetic parameters obtained from FMR measurements: effective magnetization $\mu_0 M_{\text{eff}}$, uniaxial anisotropy energy density K_u , quality factor $Q = 2K_u/(\mu_0 M_s^2) = 1 - (\mu_0 M_{\text{eff}}/\mu_0 M_s)$, Landé g factor, Gilbert damping α , and inhomogeneous broadening ΔH_{pp}^0 . The bracketed number expresses the uncertainty in the final digits.

sample	$\mu_0 M_{\text{eff}}$ (mT)	K_u (kJ/m ³)	Q	g	α	ΔH_{pp}^0 (mT)
CoFeB/Ni	222.8(1)	210(22)	0.74(4)	2.148(1)	0.0147(2)	0.09(5)
Co/Ni	-81.81(16)	473(42)	1.08(1)	2.173(27)	0.0193(25)	33.1(5)

the Fig. 7(a) inset shows exemplary raw FMR signals at $f = 30$ GHz. The error bars for the $f_{\text{res}}(H_{\text{res}})$ dependence are well below the symbol size, indicating a high accuracy of the measurement and the obtained magnetic parameters in Table II. For CoFeB/Ni, the resonance occurs at smaller fields (at $\mu_0 H_{\text{res}} \approx 0.4$ T) because of a smaller saturation field compared to Co/Ni (see Fig. 4). Generally, the linear dependence is evident in the OOP geometry and well described by Eq. (3). Moreover, from this linear relation, the quality factor Q can be calculated (see caption of Table II) by relating the uniaxial anisotropy constant K_u , determined from Eq. (4), to the shape anisotropy $K_{\text{shape}} = \frac{1}{2}\mu_0 M_s^2$, providing a convenient parameter for classifying the magnetic anisotropy of the two samples. With a Q value of 0.74, the CoFeB/Ni sample represents a weak-PMA system, where the shape anisotropy dominates over the uniaxial anisotropy. In contrast, Co/Ni, with a Q value of 1.08, exhibits stronger PMA, where the uniaxial anisotropy slightly exceeds the shape anisotropy.

In Fig. 7(b), the frequency dependency of the FMR linewidth is shown. The huge difference in the inhomogeneous broadening ΔH_{pp}^0 (intercept) is immediately apparent, with a 370-times larger value for Co/Ni compared to CoFeB/Ni (see Table II). By fitting the data to Eq. (5), one can extract the Gilbert damping parameter α from the slope, as depicted in Table II. The Gilbert damping value of $\alpha = 0.0193$ for Co/Ni lies at the lower end of the range reported in the literature [77], while CoFeB/Ni reduces this even further by approximately 25%, reaching $\alpha = 0.0147$.

IV. DISCUSSION

Initially, the use of Co₆₀Fe₂₀B₂₀ in the ML leads to a more uniform microstructure with weakened crystallographic texture and reduced crystalline coherence lengths, while preserving the overall fcc(111) growth imposed by the Pt seed layer and Ni layers in the ML. We suggest that the small amount of boron ($\approx 7\%$ atomic of the overall ML) helps to limit excessive grain growth. This results in a lower mean grain size with a narrower overall distribution, and a reduced IP coherence in the CoFeB/Ni sample. The boron is likely still present within the grains, where it may help reduce local stress and strain, thereby leading to short-range (intragranular) structural variations. However, this comes at the expense of reduced crystallographic texture, reduced OOP coherence and considerably weaker PMA, where the latter may have several causes. For example, the reduced ML texture is expected to reduce the magnetocrystalline contribution to the magnetic anisotropy [40]. Furthermore, the presence of boron and Fe near the Co/Ni interfaces may further reduce the surface and magnetoelastic anisotropy contributions, despite the use of a Co-rich alloy that would otherwise promote strain-induced

anisotropy at the Co/Ni interface. Finally, a higher degree of intermixing than in the Co/Ni sample could additionally weaken the interfacial anisotropy.

Although partial intermixing at the interfaces is commonly observed in sputter-deposited MLs, complete intermixing of the ML stack is unlikely. The absence of distinct layering in the STEM images of either sample is primarily a consequence of the low chemical contrast between the constituent elements and therefore cannot be taken as evidence of complete intermixing. The weak superstructure peak in the XRR measurements for Co/Ni, the first-order ML satellite peak (ML-1) in the XRD scans, and the significant PMA determined by SQUID-VSM and FMR measurements for both samples are difficult to reconcile with complete intermixing of the ML stacks. In particular, complete intermixing would be expected to largely suppress the interfacial PMA and favor an IP magnetic easy axis. Therefore, we conclude that the ML structure is at least partially preserved in both samples, although the degree of intermixing cannot be determined from the present data.

However, despite the weaker PMA, the short-scale structural variations in CoFeB/Ni presumably facilitate domain-wall propagation. Any deviations from the average, such as larger grains, protruding grains, broader grain-size variation, or a higher density of stacking faults and twinning, as observed in the Co/Ni sample (see Sec. III A), could serve as potential pinning sites for propagating domain walls. To correlate the distinct structural properties with the pinning potential, we estimate the characteristic domain-wall width δ_w of the samples using the following expression [76]:

$$\delta_w = \pi \sqrt{\frac{A_{\text{ex}}}{K_u}}, \quad (6)$$

where A_{ex} is the exchange stiffness, which is approximated by the composition-weighted average of Co ($A_{\text{Co}} = 31$ pJ/m), Fe ($A_{\text{Fe}} = 22$ pJ/m), and Ni ($A_{\text{Ni}} = 8$ pJ/m) [57], yielding an estimated value of $A_{\text{ex}}(\text{Co}_3\text{Ni}_6) \approx 16$ pJ/m and $A_{\text{ex}}[(\text{CoFeB})_3\text{Ni}_6] \approx 13$ pJ/m for Co/Ni and CoFeB/Ni, respectively. Using the experimentally determined values of K_u from Table II in Eq. (6) yields characteristic domain-wall widths of approximately 25 nm for CoFeB/Ni and 18 nm for Co/Ni. Owing to its lower PMA, CoFeB/Ni exhibits a slightly larger estimated domain-wall width than Co/Ni. Consequently, the mean lateral grain size determined from the STEM images (≈ 14 nm) is considerably smaller than the estimated domain-wall width (see also Fig. S2, SM [49]). As a result, domain walls extend across several grains and experience an averaged microstructural environment [78]. In contrast, the mean lateral grain size of Co/Ni (≈ 20 nm) is similar to the estimated domain-wall width.

This makes domain-wall propagation more susceptible to local variations between grains and their boundaries, thereby possibly increasing the pinning potential.

Combining this simple estimate of δ_w with the more cohesive and denser grain microstructure provides a plausible link between the experimentally observed structural and magnetic properties. For CoFeB/Ni, the smaller mean grain size together with the larger domain-wall width results in an almost vanishing remanent moment in the OOP hysteresis loop, as each domain wall interacts with a larger number of grains. In contrast, the larger mean grain size, comparable to the domain-wall width, together with the broader grain-size distribution in Co/Ni, promotes stronger grain-induced domain-wall pinning, leading to a substantially higher remanent magnetization that stabilizes a more disordered and irregular stripe-domain state. Although the two samples differ in their magnetic energy balance (different Q values), the nearly vanishing remanent moment of CoFeB/Ni ($M_r/M_s \approx 0.03\%$) cannot be attributed solely to its lower Q . For example, Co/Pt MLs with $Q < 1$ can still exhibit finite remanent magnetization, typically around $M_r/M_s \approx 1\%$ [67], suggesting that the higher microstructural uniformity and the associated weak domain-wall pinning play a decisive role in suppressing the remanent magnetization. Conversely, $Q > 1$ does not necessarily lead to large M_r/M_s [79] as observed in our Co/Ni sample, but instead is caused by the nonuniform microstructure.

The IP reversal loops are mainly governed by intrinsic magnetic differences between the samples, although the observed hysteresis can also contain contributions from the microstructure. For CoFeB/Ni, a higher remanent IP magnetization and a more abrupt switching behavior are apparent, suggesting the more efficient alignment of the stripe domains with the IP field direction (see Fig. 4(c), second column) and less hindered switching of the Bloch-type domain-wall centers [65]. The highly aligned remanent stripe-domain state already observed after IP saturation is consistent with its weak-PMA regime, where the reduced energy difference between IP and OOP magnetization states enables the formation of elongated stripe domains while allowing efficient domain rearrangement. This emphasizes the importance of balancing magnetic anisotropy but also tuning the microstructure to stabilize such a well-ordered stripe-domain state. In contrast, caused by the underlying defect-rich microstructure, the Co/Ni sample retains only a weakly aligned stripe-domain state after IP demagnetization, with intermittent stripes (see Fig. 4(d), second column). The more gradual reversal observed for Co/Ni in the OOP loop is likewise reflected in the IP loop. However, this should not be attributed alone to microstructural effects, as the domains are not aligned with the applied field direction, meaning that the IP field is not fully parallel to the domain walls and therefore also contributes to the gradual reversal.

Remarkably, the nearly identical domain periods observed after OOP saturation and OOP demagnetization for the CoFeB/Ni sample show that the characteristic domain length scale is reproducible and close to its equilibrium value, exhibiting only a weak dependence on the magnetic field history. On the contrary, Co/Ni exhibits different domain periods after the two field treatments and a smaller domain-period ratio of 0.70 (compared to 0.77 for CoFeB/Ni) between the IP- and

OOP-demagnetized states, indicating a stronger influence of the microstructural disorder and domain-wall pinning on the domain configuration [80].

The role of disorder is further reflected in the RPM measurements, where Co/Ni exhibits higher cross-correlation coefficients throughout minor-loop field cycling compared to CoFeB/Ni, as expected for systems with stronger disorder and pinning [71]. Conversely, CoFeB/Ni shows a steeper decrease and a vanishing RPM after minor loop and OOP saturation-field cycling, respectively, indicating a lack of reproducible domain configurations and the presence of multiple possible reversal pathways. The latter observation is further justified by the asymmetry between the dense bubble domain state observed before saturation and the subsequent nucleation of only a few stripe domains after saturation (see Fig. 5), suggesting that the initial bubble domains do not act as persistent nucleation centers [60,61].

Turning to the dynamic magnetic properties, the more heterogeneous microstructure of Co/Ni is reflected in a substantially larger inhomogeneous linewidth broadening compared to CoFeB/Ni, indicating stronger spatial variations in magnetization, magnetic anisotropy, and microstructure [81,82]. As illustrated by the exemplary FMR raw signals in the inset of Fig. 7(a), the broader shape of the Co/Ni sample's resonance profile stems from its enhanced magnetic inhomogeneity. For CoFeB/Ni, a narrower profile is apparent, reflecting a more homogeneous magnetic environment. The higher PMA of Co/Ni may also contribute to the increased linewidth broadening [83], but the dominant contribution arises from the distinct differences in microstructure.

The determined magnetic Gilbert damping values for our samples are comparable to previous studies on Co/Ni MLs (e.g., Fig. 3 and Table 1 in [77]) and thin CoFeB layers with PMA [84]. However, similarly low damping parameters ($\alpha \approx 0.02$) have also been reported for Co/Pt and Co/Pd MLs [85–87], indicating that avoiding heavy metals in the ML alone does not guarantee a reduction in magnetic damping. For example, reduced intermixing at Co/Pt interfaces has been proposed to reduce magnetic damping by modifying spin-pumping contributions [86]. We emphasize that the magnetic damping parameter obtained from FMR includes contributions from both intrinsic and spin-pumping, which makes separating them individually difficult. Accordingly, the observed difference in magnetic damping of approximately 25% between our samples may arise from multiple factors, including spin-pumping and intrinsic contributions. Based on the performed experiments, the relative contributions of the individual mechanisms cannot be determined, and the following discussion remains qualitative and preliminary, outlining several plausible explanations.

Despite possible spin-pumping contributions stemming from the ML interfaces, spin-pumping into the Pt seed and cap layer surrounding the ML contributes to the reported magnetic damping as well. This has been demonstrated, for example, in permalloy thin films [88]. Furthermore, since the first and last magnetic layers in our samples are either CoFeB or Co, differences in the CoFeB/Pt and Co/Pt spin-mixing conductance could modify the spin-pumping from the Pt seed and cap layer and may contribute to the observed difference in α [89–91]. To separate such effects, a systematic thickness-dependent study

varying the repetition number, seed, and cap layer materials, and their thicknesses while preserving the ML structure would be required, as the influence of interface-related contributions increases for thinner magnetic layers [92]. Alternatively, replacing Pt with fcc(111)-promoting materials with lower spin-pumping efficiencies, such as Cu, Au, or Ti, could further reduce interface-related contributions [12,27,39,93,94]. However, such an investigation is beyond the scope of the present work. In fact, spin-pumping contributions to the magnetic damping from the Pt seed and cap layer are at least reduced in the samples, given the high number of repetitions [89,92].

Finally, intrinsic differences between the MLs may contribute to the reduced magnetic damping upon replacing Co with CoFeB. The higher PMA and g factor of Co/Ni, reflecting stronger spin-orbit coupling-related effects, may also contribute to its higher magnetic damping [20], while reported differences between Co and CoFeB thin films suggest intrinsic material-related effects as another possible origin [26,82,95–97]. However, separating intrinsic contributions remains challenging in thin-film systems, as the measured magnetic damping is strongly influenced by the adjacent layers, as noted above.

While the microscopic origin of the reduced damping cannot be conclusively identified from the present experiments, we favor either different spin-pumping and/or an intrinsic contribution associated with the CoFeB/Ni ML. Consequently, the reported values should be regarded as an upper bound on the magnetic Gilbert damping, motivating further studies to distinguish between spin-pumping and intrinsic damping mechanisms.

Applicationwise, the distinct microstructural and magnetic characteristics of the two MLs make them attractive for different application scenarios. The more homogeneous microstructure, reduced domain-wall pinning, lower inhomogeneous linewidth broadening, and lower Gilbert damping of CoFeB/Ni are advantageous for applications requiring efficient and reproducible magnetization dynamics [47], such as magnonic devices [98,99], spin-transfer- and spin-orbit-torque devices [100,101], fast domain-wall-based technologies [28,102], and neuromorphic computing [103]. Furthermore, weak stripe-domain states have been extensively investigated in PMA systems, where they arise from the competition between exchange interactions, anisotropy, and dipolar energy. Such configurations have been reported in various materials, including FePt(Pd) [104,105], Co/Pt [69,106], FeN [107], and Fe/Co [108,109] MLs or thin films. The highly aligned stripe-domain states observed for CoFeB/Ni after IP field treatment can be considered as part of this broader class of weak stripe-domain configurations. In contrast, the stronger pinning, enhanced RPM, and higher remanent magnetization of Co/Ni provide a more stable magnetic domain configuration, which may be advantageous for applications requiring robust domain states, such as magnetic memory and patterned magnetic structures. However, the comparatively low magnetic Gilbert damping, combined with the higher PMA, makes the Co/Ni ML equally suitable for the aforementioned spintronic applications.

To achieve a microstructure in Co/Ni comparable to that of the CoFeB/Ni sample, different approaches could be con-

sidered. For our sputtering conditions, the sputtered atoms are expected to be fully thermalized [110,111]. Consequently, the primary source of energy delivered to the growing film is the bombardment by sputter-gas atoms, like neutralized process gas atoms [32,110,112,113]. At the deposition pressure used in this work, the transferred energy is likely insufficient to promote significant surface diffusion and atomic rearrangement, preventing the formation of a smoother grain microstructure in the Co/Ni sample. Achieving such a microstructure requires additional energy to increase surface mobility after condensation on the substrate. This may be accomplished, for example, by reducing the deposition pressure to increase the kinetic energy of the arriving species or by applying *in situ* substrate heating during deposition [19,93]. Postdeposition treatments such as annealing [35,114–116] are likely not suitable since it only modifies an already formed microstructure. For CoFeB/Ni, postdeposition annealing could increase the PMA through boron redistribution, thereby increasing interfacial anisotropy, while preserving the short-range structural order as long as the annealing temperature remains below the onset of bulk diffusion. More generally, transition-metal alloys containing elements such as boron or zirconium [117,118], may provide a promising route toward smoother or even amorphous microstructures while preserving or enhancing PMA in the as-deposited state.

V. CONCLUSION

We have investigated the structure, together with the static and dynamic magnetic properties of polycrystalline sputter-deposited [CoFeB/Ni]₅₀ and [Co/Ni]₅₀ MLs. Our main finding is that replacing Co with CoFeB results in a more uniform microstructure with smaller grains and a narrower size distribution, albeit at the expense of reduced crystallographic texture, crystalline coherence and weaker PMA. The smoother microstructure gives rise to remanent magnetic states exhibiting near-complete magnetic compensation, as well as highly periodic isotropic labyrinth and well-aligned stripe-domain configurations after simple OOP and IP saturation, respectively. Furthermore, CoFeB/Ni exhibits a 25% reduction in the magnetic Gilbert damping parameter ($\alpha = 0.0147$) compared to Co/Ni ($\alpha = 0.0193$), accompanied by an approximately 370-fold reduction in the inhomogeneous linewidth broadening, highlighting its potential for low-power perpendicular spintronic and magnonic devices, particularly in applications requiring controlled stripe and bubble domain states. Further studies could clarify whether the observed reduction in α arises from variations in spin-pumping contributions or intrinsic damping mechanisms. Additionally, it remains of interest whether the CoFeB/Ni or Co/Ni systems can be further optimized to combine their respective advantages while preserving the low-pinning microstructure.

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DATA AVAILABILITY

The data that support the findings of this article are openly available [72].

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