



TECHNICAL ARTICLE

Engineering Superior Soft Magnetic Properties Through FeCo/Cr and FeCo/Fe Nanocrystalline Architectures

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Soft magnetic materials are integral to various technologies, including high-frequency transformers, spintronic devices, and magnetic sensors, because of their low coercivity (H_C) and high magnetization. This study investigates nanocrystalline $\text{Fe}_{70}\text{Co}_{30}(100\text{ nm})/\text{Cr}(2\text{--}10\text{ nm})$ and $\text{Fe}_{70}\text{Co}_{30}(100\text{ nm})/\text{Fe}(2\text{--}10\text{ nm})$ thin films, deposited via magnetron sputtering. X-ray diffraction analysis reveals that $\text{Fe}_{70}\text{Co}_{30}$ films without buffer layers crystallize in a body-centered cubic phase. The Cr and Fe buffer layers enhance it by minimizing lattice mismatch and promoting a uniform crystalline growth of the $\text{Fe}_{70}\text{Co}_{30}$ layer. The most significant particle size variations observed were approximately 24% for the Cr buffer and 31% for the Fe buffer. Microstructural characterization using atomic force microscopy and scanning electron microscopy identified nanograins with an average size of 30–80 nm. Magnetic characterization with a vibrating sample magnetometer showed a marked reduction in H_C . Additionally, the remanence-to-saturation magnetization ratio remains nearly constant at 0.95 across all Fe-buffered films, whereas it increases from 0.80 to 0.92 in Cr-buffered films as the Cr thickness increases. The results highlight the importance of buffer layer engineering in enhancing the microstructural and magnetic properties of FeCo thin films, thereby increasing their suitability for advanced magnetic technologies.

INTRODUCTION

Soft magnetic thin films are increasingly important for next-generation electrical and power systems that require miniaturization, fast switching, and reduced energy consumption. Recent studies have demonstrated that functional magnetic materials featuring controlled nanocrystalline microstructures are crucial elements in compact power converters, electromagnetic interference filters, magnetic logic units, and high-frequency inductors directly integrated onto semiconductor platforms.^{1–5} With the increasing need for wearable and flexible electronics, there is a corresponding interest in magnetic materials characterized by low coercivity (H_C), high saturation magnetisation (M_S), and magnetic softness that is unaffected by bending or stretching.⁶ In advanced composite architectures,

soft magnetic layers have been shown to enhance electromagnetic performance, mechanical robustness, and energy dissipation pathways, making them valuable in both structural electronics and multifunctional devices.⁷ Functional magnetic thin films are increasingly integrated into high-density data storage systems and spintronic devices, alongside traditional power conversion functions. In these applications, precise control of grain size, interfacial strain, and magnetic anisotropy directly influences switching efficiency and magnetoresistive properties.⁸ These improvements show the importance of scalable, affordable FeCo-based thin films with microstructural and magnetic properties that can be changed by choosing the right buffer layer and controlling the deposition settings. The goals of the studies are to improve understanding of basic properties and to make materials that are faster, longer-lasting, and use less energy.^{9–15} Another important contribution of CoFe to magnetism is the wide range of applications enabled by

its combination with rare-earth elements,¹⁶ underscoring the heightened scientific and technological significance of CoFe-based materials.

Within this broad range of applications, specific magnetic performance metrics become critically important. Achieving low H_C and high magnetic induction ($B_S = 4\pi M_S$) in magnetic thin films is of great interest because of their potential applications in high-frequency transformers, magnetic sensors, and spin electronic devices. For instance, magnetic random-access memory using spin-transfer torque (STT) shows significant potential as a next-generation memory technology that meets these demanding requirements.¹⁷ Another essential requirement is a high remanent magnetization (M_R) value in the hysteresis loop of magnetic thin film. A writing head core material with a soft magnetic buffer layer for perpendicular double-layered media requires a soft magnetic film with a low H_C and high remanence/saturation magnetization ratio (M_R/M_S) to record data with a high density. Incorporating in-plane magnetic anisotropic materials with low H_C and high B_S into the memory effectively increases memory density and write capabilities through the STT effect.^{18,19} However, the use of rare or exotic elements or consecutive complex operations in these in-plane magnetic anisotropy materials often results in high costs for these applications.^{15,20–26} Therefore, including materials with lower-budget components in such applications could potentially alter the course of these investigations.

Scientists are actively investigating methods to enhance current soft magnetic technologies, driven by the increasing need for effective energy conversion, storage, and data processing. In this effort, Fe-Co-based magnetic thin films have emerged as an important material platform. Due to their great thermal stability, cost-effectiveness, low H_C , and high M_S , Fe-Co-based films are extensively employed in a variety of single- and multilayer structures, and the films are produced under optimal conditions.^{24,25,27–37} The Fe₇₀Co₃₀ system, with a composition that includes a B_S value of up to 2.4 T and H_C , 100 Oe, stands out among these films. However, to further improve its performance, introducing buffer layers between the Fe-Co layer and the substrate can play a crucial role. These buffer layers can modify the interface structure, enhance the crystallographic texture, and mitigate interdiffusion between the Fe-Co layer and the substrate.

Building upon the advantages of Fe-Co-based magnetic thin films, researchers have turned their attention to the role of buffer layers in fine-tuning the properties. By introducing buffer layers between the Fe-Co layer and the substrate, it is possible to achieve even greater control over factors such as crystallographic orientation and interfacial interactions, ultimately leading to improved soft magnetic behavior.^{15,20–22,24,27–29,35,38} Fu conducted a study on the magnetic properties with a composition of Fe₆₅Co₃₅. The study explored the effects of various

substrates and advocated the use of buffer layers to change the structure of the films as a vital strategy for improving their soft magnetic characteristics.³⁹ Through systematic alterations in composition, thickness, and deposition procedures, researchers have optimized the characteristics and performance of these films, paving the way for new applications. Fu shows that the Co buffer layer significantly affects the crystallographic microstructure and grain size of the Fe-Co, and the primary contributor to the soft magnetic behavior can cause a change in crystalline size.²⁸ With high B_S across different film thicknesses, Shintaku et al. sought to achieve ideal soft magnetic characteristics in Fe-Co system films by investigating the influence of different additives and buffer layers. Al₂O₃, which precipitated out of Fe-Co films at grain boundaries, formed a continuous columnar structure, greatly lowering their H_C . It plays a crucial role in facilitating the formation of small grains and random crystal orientation, which are necessary to achieve desirable soft magnetic properties.³⁰ Using a Cu buffer layer significantly changed the M-H loop and H_C , as shown by Wang et al.²² The Cu buffer layer's considerable roughness and crystalline size likely contributed to the H_C 's increase from 2 Oe in the absence of Cu to 37 Oe in the presence of it. Pakale et al. illuminate the exchange bias field (H_{ex}) sensitivity to interface roughness, crystalline structure, and grain size in ferromagnetic-antiferromagnetic film systems, and the results show that magnetostatic influences on domain creation cause H_{ex} to rise beyond critical buffer-layer thicknesses, with top and bottom configurations having different roughness dependencies.^{21,40} Another study examined how the non-magnetic Ta buffer and cap layers affect the Co-Fe film microstructure and magnetic characteristics at different temperatures, and H_C increases dramatically with the Ta buffer and cap layers and with annealing temperature.⁴¹ Based on our current understanding, annealing might result in incompatibilities and defects between the layers, making it a process that is not highly recommended unless absolutely essential. Researchers have reported that soft magnetic buffer layers may diminish Co₉₀Fe₁₀ thin film's H_C , particularly when grown in a magnetic field.³⁴ An investigation into the impact of Fe-Co/Co thin film revealed an enhancement in value up to $B_S = 24$ kG and down to $H_C = 3$ Oe. Fu et al. analyzed a decrease in average grain size from 74 nm to approximately 8.2 nm, likely attributed to this improvement.³⁵ Interdiffusion and considerable magnetic moment loss occur at the interfaces of multilayer structured films due to lattice mismatch, surface and interface fluctuations, and so on, which incur extra expenses for recovery.³⁸ The structure and magnetic properties of texture-structured Fe-Co films on Ta/Pd seed layers were better than those on glass substrates or Ta seed layers. A 15% Fe content was found to be best for fcc phase formation.²⁶ Liu et al. investigated the

impact of Co buffer layers on soft magnetic FeCo films. They found that a 2-nm Co substrate significantly reduced the average grain size from 74 nm to 8.2 nm. Additionally, the Co substrate increased the strain by a factor of approximately two. Despite this increase in strain, the coercive field, H_C , was notably reduced from nearly 100 Oe to 3 Oe, respectively.⁴²

Considering the benefits of Fe-Co-based thin films, one can see that buffer layers play a crucial role in their refinement. By selectively adding buffer layers between the Fe-Co layer and the substrate, researchers hope to precisely control the microstructural and magnetic properties of thin films. This will open new ways to improve their performance in a wide range of applications. Existing research lacks a thorough study of a buffer layer that requires no further processing. Although numerous studies have shown that seed and buffer layers (Co, Cu, Pd/Pt, Cr, etc.) modify FeCo texture,^{14,27,34,42–47} grain size, and H_C , a direct, side-by-side quantitative comparison of a ferromagnetic (Fe) and a Cr buffer series—performed under identical deposition conditions and across the same buffer-thickness window (2–10 nm)—remains lacking. The present work fills this gap by correlating thickness-dependent structural metrics, morphology, density/roughness, and magnetic performance in as-deposited films by systematically varying the thicknesses of the buffer layers; the structure could be optimized to achieve the desired soft magnetic behaviors.

EXPERIMENTAL

Polycrystalline Fe₇₀Co₃₀ thin films were deposited by magnetron sputtering from an elemental target, with the substrate rotated at 12 rpm in a single-chamber system at about ambient temperature. The elemental mapping and energy dispersive x-ray analysis of the target are shown in the supplementary information (Fig. S1). The Fe₇₀Co₃₀ system, the primary focus of the research, is denoted as FeCo for simplicity. The sputtering process was conducted at a base pressure $> 1.5 \times 10^{-6}$ Torr before deposition, using different Cr or Fe thicknesses on SiO₂/Si(100) substrates. A working gas pressure of 5 mTorr and a sputtering power of 30 W were constantly maintained throughout the deposition process. The 2-nm Si capping layer was deposited using 50 W sputter power at 5 mTorr Ar pressure. The series of stack structures consisted of Si(100) substrate, followed by either Cr (2-5-8-10 nm) or Fe (2-5-8-10 nm), then FeCo (100 nm). A 2-nm Si capping layer was deposited using a sputter power of 50 W at a 5 mTorr Ar pressure.

X-ray diffraction (XRD) measurements were performed using the Rigaku SmartLab diffractometer with CuK $_{\alpha}$ radiation ($\lambda = 1.5354$ Å). Rietveld refinement was employed for profile matching and to refine the unit cell parameters.⁴⁸ Instrumental broadening was corrected using a standard Si

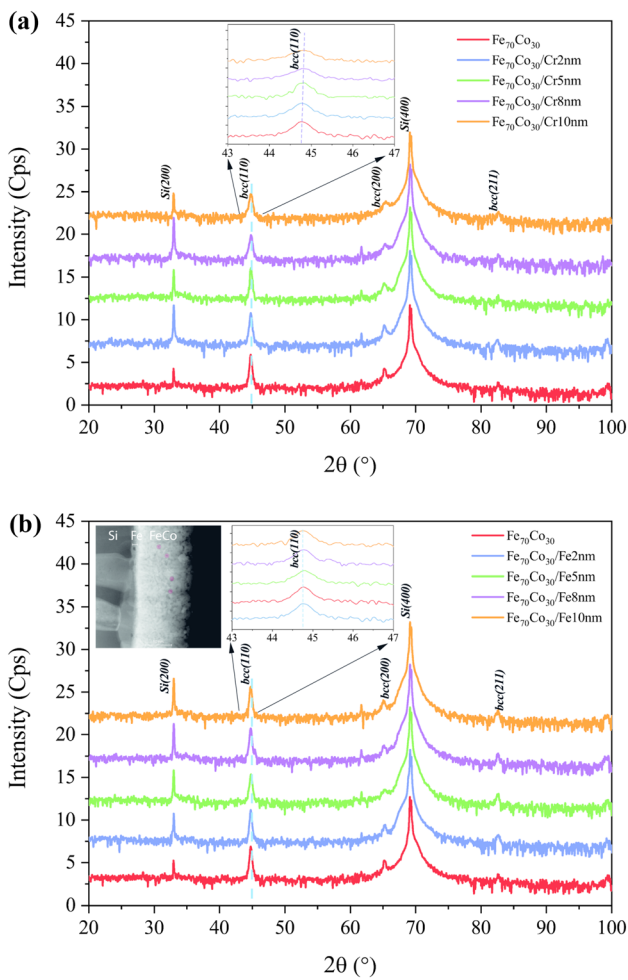
sample with a known large crystalline size, ensuring the accurate determination of crystalline size by the Debye-Scherrer (D-S) equation. Additionally, residual stresses and microstrain were accounted for using the Williamson-Hall (W-H) plot, allowing separation of broadening contributions from crystalline size and microstrain.⁴⁹ Crystalline size does not refer to individual atoms or molecules but rather to coherent domains within the polycrystalline film. The lattice constant and unit cell volume of thin films are computed and shown in Table I as per the supplementary material. Using both methods is crucial for the present study, as the films may undergo strain, which can broaden XRD peaks. Relying solely on a single method could lead to inaccurate estimates of crystallite size. By comparing the results from both methods, more reliable conclusions can be ensured, particularly for strained thin films, where strain can significantly affect crystallite size calculations. Also, the thickness of the thin films and buffer layer was determined using x-ray reflectance (XRR) data before growth, and calibration data were created and applied (Rigaku SmartLab). A quartz crystal oscillator positioned next to the sample monitored film thickness during deposition. The composition and homogeneity measurements were performed using scanning electron microscopy (SEM, Jeol). SEM micro images of thin films are given in the supplementary information (Fig. S2). For all films, energy-dispersive x-ray spectroscopy (EDX) measurements were performed. The detailed microstructural measurements were carried out using a transmission electron microscope (TEM, Thermo Fisher Scientific). To reduce the impact of shape anisotropy, all films were deposited on square substrates (as seen in Fig. S3). An atomic force microscope (PARK XE100 AFM) was used to gather pictures of the surface morphology, and the particle size distribution was examined using its software. The hysteresis loops of all thin films were measured at room temperature using a vibrating sample magnetometer (Cryogenics CFMS 9 T-VSM), with the magnetic field varied up to ± 6 kOe to saturate the magnetic moments, where the M_S was determined from the M-H curve plateau, and H_C was identified as the field strength at which the net magnetization crossed the zero axis during field reversal.

RESULTS AND DISCUSSION

Figure 1 shows the XRD patterns of the thin films that were deposited on Cr or Fe buffer layers. The FeCo thin film, without a buffer layer, reveals that the wide peaks are due to the existence of the FeCo body-centered-cubic (bcc) with space group number 229 (Im3m), and the lattice parameter $a = 2.8605$ Å is in line with the literature.²³ The three typical diffraction planes with 2θ values of 44.658°, 66.870°, and 82.569° correspond to the (110), (200), and (211) Miller indices of the crystalline Fe-Co alloy JCPDS

Table I. The calculated lattice parameter, unit cell volume, crystalline size from Debye-Scheerer, crystalline size from Williams-Hall, variation between D-S and W-H, and ratio of I_{211}/I_{110} Bragg peaks

System	Lattice parameter (nm)	Unit cell volume (nm ³)	Crystalline size from (D-S) (nm)	Crystalline size from W-H (nm)	Lattice strain (ϵ)	Variation in D-S and W-S (%)
FeCo	2.860	23.393	21	19	0.04457	8.16
Cr2nm	2.859	23.376	18	16	0.05135	8.16
Cr5nm	2.858	23.354	18	16	0.05143	8.23
Cr8nm	2.857	23.332	17	15	0.05449	8.16
Cr10nm	2.856	23.307	15	14	0.05874	8.18
Fe2nm	2.861	23.418	16	15	0.05740	8.20
Fe5nm	2.861	23.423	15	14	0.06120	8.18
Fe8nm	2.861	23.435	14	13	0.06271	8.16
Fe10nm	2.862	23.437	14	13	0.06509	8.25

**Fig. 1.** XRD patterns plotted on a logarithmic scale of samples (a) FeCo/Cr(2-10 nm) and (b) FeCo/Fe(2-10 nm) inset of zoom in FeCo bcc (011) Bragg position and TEM image of Fe 10-nm-thin film. The (110) plane diffraction peaks are prominent.

49-1567, respectively. Bcc (110), (200), and (211) peaks were observed for all the samples. The cell parameters of FeCo with different buffer layers were computed and are shown in Table I, consistent

with the values provided in Refs. 29 and 39. The inset in Fig. 1b shows a slight shift of the bcc (110) peak toward lower 2θ values with increasing Fe buffer layer thickness, suggesting possible strain relaxation or a modification in crystallographic texture.⁵⁰ In contrast, the peak position remains nearly constant with increasing Cr buffer layer thickness, implying a more stable lattice structure. Localized alterations, such as defects, may transpire without influencing peak positions. The fact that the unit cell parameters of the FeCo layer exhibit a change with the Cr buffer layer suggests a higher level of residual strain compared with the Fe buffer layer. This stability is critical, as it indicates that the Cr buffer layer can fulfill its role without significantly altering the main layer's strain properties, which could help maintain the desired magnetic characteristics. The consistent presence of peaks from the three primary diffraction planes of the FeCo phase across all samples further confirms that both Cr and Fe buffer layers effectively support the growth of the FeCo phase without introducing significant structural distortions. Also, TEM imaging (Fig. 1b inset) of the 10-nm Fe-buffered sample confirms that the Fe/FeCo interface is abrupt, with no observable interdiffusion or intermixing.

The correction of this ratio using the background signal during data analysis was crucial for accurate determination of the relationship between these planes. Notably, the ratio increases dramatically in films with buffer layers of 8–10 nm thickness, approaching a saturation threshold. This behavior is consistent with previous studies that observed similar trends in FeCo films grown on different substrates.³⁰ Using different buffer layers altered the crystal orientation of FeCo system films, as evidenced by the expansion of the strain.

The quantitative study assessed the phase composition, lattice parameter, crystalline size, and degree of crystallinity of the thin films. The crystallite size was determined using two complementary approaches: D-S and W-H, with detailed procedures provided in the Supplementary Material. Table I presents the crystalline size of films

determined using the D-S and W-H techniques. The crystalline size distributions obtained with the W-H approach were typically smaller than those obtained with the D-S method, and this variation was due to differences in how the crystalline size distribution was averaged. It was noted that the crystallite sizes obtained via D-S and W-H analyses differ by a small but nearly constant offset across all samples. Numerically, the percent difference $\Delta = (D_{DS} - D_{WH})/D_{DS} \times 100\%$ has a mean value of 8.6% with a population standard deviation of 2.1% (range 6.3–11.8%) for the nine samples. This behavior indicates a consistent strain-size decoupling, in which the microstrain contribution to peak broadening remains relatively uniform across varying buffer layer thicknesses. While the W-H method accounts for strain-induced broadening, the D-S approach reflects primarily size-related effects, leading to a systematic offset between the two estimates. In FeCo thin films, such uniform microstrain is physically reasonable and suggests that buffer-layer-induced structural modifications mainly influence texture evolution rather than inducing significant grain growth. Furthermore, in physical terms, this indicates that the films retain nanocrystalline coherent domains with low-to-moderate microstrain. The strain evolves smoothly with buffer thickness and does not introduce abrupt lattice distortions that would cause large, sample-dependent deviations among different analysis methods. The growth of FeCo without buffer layers results in the formation of grains with very small nuclei, consistent with previous reports.²⁸ The findings indicate that both the Fe and Cr buffer layers serve as agents that improve the grain structure. Additionally, small FeCo grains form on the Fe or Cr layer and grow upward. The observed variation in H_C with grain size aligns with the random anisotropy hypothesis for nanocrystalline films.⁵¹ Lattice strain increases with Fe and Cr buffer layer thickness in Fe₇₀Co₃₀ films because of lattice mismatch, which accumulates as the buffer grows thicker. This constraint causes greater distortion and strain energy in the film, especially when there are significant differences in thermal expansion or lattice constants. Beyond a critical thickness, defects may form to relieve the strain, emphasizing the need for precise layer engineering.

The W-H analysis reveals a systematic increase in ε with buffer layer thickness for both the Fe and Cr series. The analysis results are given in Table I. This monotonic increase suggests that strain accumulates gradually because of lattice mismatch between FeCo and the underlying layers. Comparable strain evolution has been observed in other nanocrystalline metallic and oxide thin-film systems, where changes in microstrain are attributed to lattice distortion arising from interface matching and the progressive development of grain boundaries as the film thickness increases.⁵² In the present FeCo thin films, the measured microstrain

values (0.0446–0.0651) indicate a low-to-moderate distortion regime in which coherent lattice regions remain stable but are increasingly influenced by interfacial strain fields. As demonstrated in the referenced study,⁵³ variations in microstrain affect the grain-growth pathway by modifying lattice distortion at grain boundaries, thereby influencing the rate and uniformity of crystallite enlargement. Rather than promoting abrupt coalescence or abnormal grain growth, the presence of moderate microstrain supports a more gradual and structurally coherent evolution of nanocrystalline domains. This behavior is consistent with our AFM and SEM observations, which show progressive grain enlargement in both Fe- and Cr-buffered films without any sudden coarsening transitions. Furthermore, the microstrain trend correlates with the magnetic response: changes in lattice distortion influence the effective anisotropy landscape and domain-wall mobility, contributing to the observed reduction in H_C as buffer-layer thickness increases. Taken together, the W-H analysis provides not only a quantitative measure of strain but also valuable insight into how strain relaxation and grain boundary evolution collectively enhance the soft magnetic performance of FeCo thin films.

The surface topography of the thin films was characterized using AFM measurements, as seen in Fig. 2a–j. Additional detailed experimental data, including amplitude and phase force microscopy images, are provided in the supplementary information (Figs. S4 and S5a–j). These images clearly show an increase in root-mean-square surface (rms) roughness, which can be estimated to be in the range of 1.5–3 nm as the thickness of Cr or Fe increases. The increase in roughness is consistent with the reduction of lattice strain observed in the FeCo phase using XRD. Buffer layer incorporation increases grain size in the films, and nanocluster formation is indicated in Cr or Fe buffer layers. SEM measurements detected these particles, which are provided in the supplementary information (in Fig. S2). Films grown on a buffer layer exhibit distinct microscopic characteristics, such as nanoclusters and increased rms roughness, which affect the films' magnetization. In contrast, films without buffer layers displayed a more uniform particle distribution over a smooth surface. The introduction of Fe and Cr buffer layers, however, dramatically altered this scenario. Specifically, the Fe buffer layer induced the formation of spherical particles, whereas the Cr buffer layer promoted the growth of nanorods. In both cases, the surface roughness values and particle sizes increased significantly, which may have influenced the films' magnetic properties.

Nanorods can lead to denser films by promoting better packing and reducing voids within the film structure. During growth, nanorods provide a template that encourages adjacent material to align and fill gaps more efficiently, minimizing porosity. Their

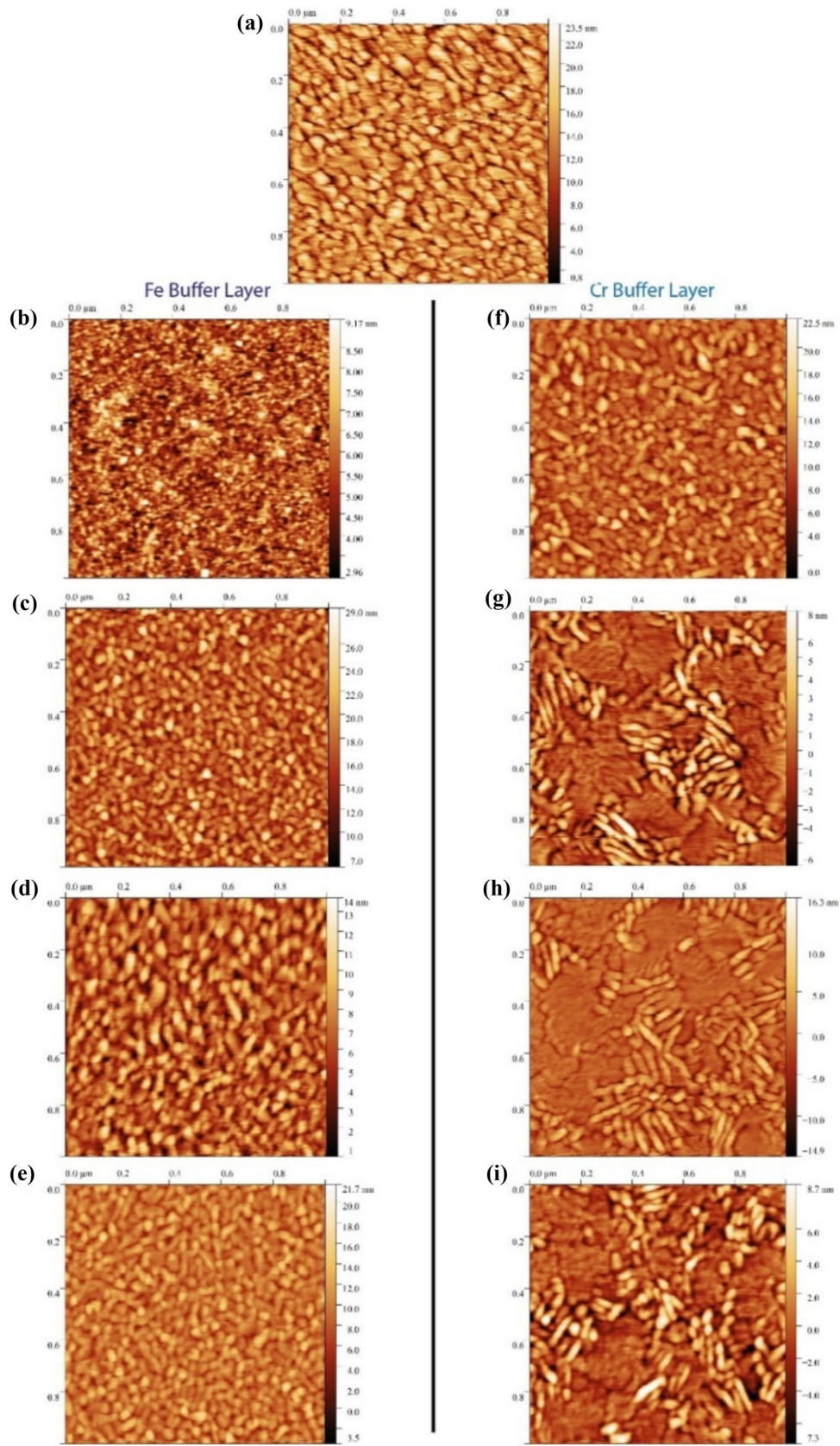


Fig. 2. AFM images of thin films: (a) FeCo, (b) FeCo/Fe (2 nm), (c) FeCo/Fe (5 nm), (d) FeCo/Fe (8 nm), (e) FeCo/Fe (10 nm), (f) FeCo/Cr (2 nm), (g) FeCo/Cr (5 nm), (h) FeCo/Cr (8 nm), and (i) FeCo/Cr (10 nm).

high surface area enhances nucleation and allows for uniform material deposition around the rods. Additionally, the vertical alignment of nanorods ensures more controlled material distribution, resulting in a compact, tightly packed film. This mechanism improves the overall density of the film compared to structures with random or uneven growth.

The particle size analyses reveal distinct morphological differences between the FeCo thin films grown on Fe and Cr buffer layers. On the Fe buffer layer, the FeCo particles exhibit a spherical morphology with an average size of approximately 50 nm. In contrast, the FeCo particles on the Cr buffer layer form nanorods, with dimensions around 30 nm in width and 80 nm in length. This variation in particle size and morphology is primarily driven by the different grain growth dynamics introduced by the Fe and Cr substrates. The Cr buffer layer, in particular, promotes enhanced grain growth, leading to a significant increase in the grain growth rate and resulting in the formation of nanorod microstructures in the thin films. These nanorod structures contribute to a denser thin film and improve the magnetic properties, as evidenced by an increase in the M_S that approaches the theoretical maximum for FeCo.⁵⁴ The ability to precisely control the microstructure and magnetic characteristics of FeCo thin films through the selection of appropriate buffer layers is a significant advantage. This control is crucial not only for applications in magnetic recording head technologies and spintronic devices but also for other magnetic devices, such as inductors, transformers, and high-frequency noise filters, which rely on the tailored magnetic properties of thin films.³⁰

Figure 3a–j displays the in-plane remanence ratio versus magnetic field hysteresis curves of single and multilayer structured films with different thicknesses for Cr and Fe buffer layers. Additionally, to enhance comprehension, the internal graphs of each thin film provide MH loops for low field values. All films exhibited soft magnetic properties. Both Fe and Cr buffer layers of 2 nm thickness are used to improve the soft magnetic properties of FeCo films, and these properties improve as the thickness increases. In addition, the graphs demonstrate that the film without a buffer layer is distinct from the other films. The hysteresis loops exhibit a change from an S-shape to a square shape. This phenomenon is supported by both M_R/M_S and H_C numerical data. Except for FeCo thin films without a buffer layer, the application of thin buffer layers ranging from 2 to 10 nm significantly lowered the H_C . By lowering the H_C to 4 Oe and raising the M_R/M_S value to 0.95, improved systems were achieved. The insertion of Cr or Fe buffer layers and optimization of their thickness are necessary to achieve excellent soft magnetic characteristics. The observation of reduced H_C in FeCo/Cr films compared with FeCo/Fe is consistent with previous work,

which attributed similar behavior to the enhanced magnetic softness due to buffer layer effects.^{15,22} However, the present results show a more pronounced reduction, which may be attributed to the specific thickness range of the Cr or Fe buffer layers used in the study.

Interdiffusion at metallic interfaces depends strongly on deposition temperature, adatom mobility, and thermodynamic drivers such as enthalpy of mixing. In our study, all layers were deposited at room temperature, where self-diffusion coefficients for Fe and Cr in bcc structures are on the order of 10^{-25} – 10^{-27} m²/s, implying diffusion lengths of only a few Å during the deposition time.^{55–57} Thus, long-range alloying or intermetallic formation is unlikely. This interpretation is supported by our structural measurements. XRD patterns show no shifts or extra reflections associated with Fe–Cr or Fe–Co–Cr intermetallic. W-H analysis yields microstrain values consistent with coherent FeCo grains rather than compositionally graded layers, and XRR fits indicate well-defined interfaces. At the same time, sputtered films typically have an interfacial mixing width of 0.2–1.0 nm due to ballistic collisions during growth, particularly at FeCo/Cr interfaces where the heat of mixing is moderately negative. Such localized chemical grading could slightly perturb exchange stiffness and domain-wall energy, providing an additional contribution to the variation of H_C and M_R/M_S observed between Fe- and Cr-buffered series.

The study on FeCo/Cr and FeCo/Fe multilayer films provides new insights into how buffer layer thickness influences H_C and M_S . The results indicate that thinner Cr buffer layers lead to a significant reduction in H_C . However, the study extends this understanding by demonstrating that the reduction in H_C is more pronounced in FeCo/Cr films than in FeCo/Fe, likely due to differences in lattice expansion and strain effects. This improved understanding of buffer layer effects has important implications for the design of high-performance magnetic materials for applications such as magnetic sensors and data storage devices.^{46,58–60} Future research should explore the effects of additional buffer layer materials and their interactions with FeCo alloys to further optimize magnetic properties. The variations in H_C and M_S with different buffer layers reveal a strong dependence on the microstructural changes induced by these layers. Specifically, the Cr or Fe buffer layers led to a notable reduction in particle size, which influenced the observed magnetic behavior. Despite ongoing efforts to enhance energy conversion and data processing technologies, the crucial role of buffer layers in improving FeCo-based magnetic thin films has often been overlooked. The study provides strong evidence for their importance, as changes in buffer layer thickness were shown to significantly affect magnetic properties, with even small adjustments leading to measurable

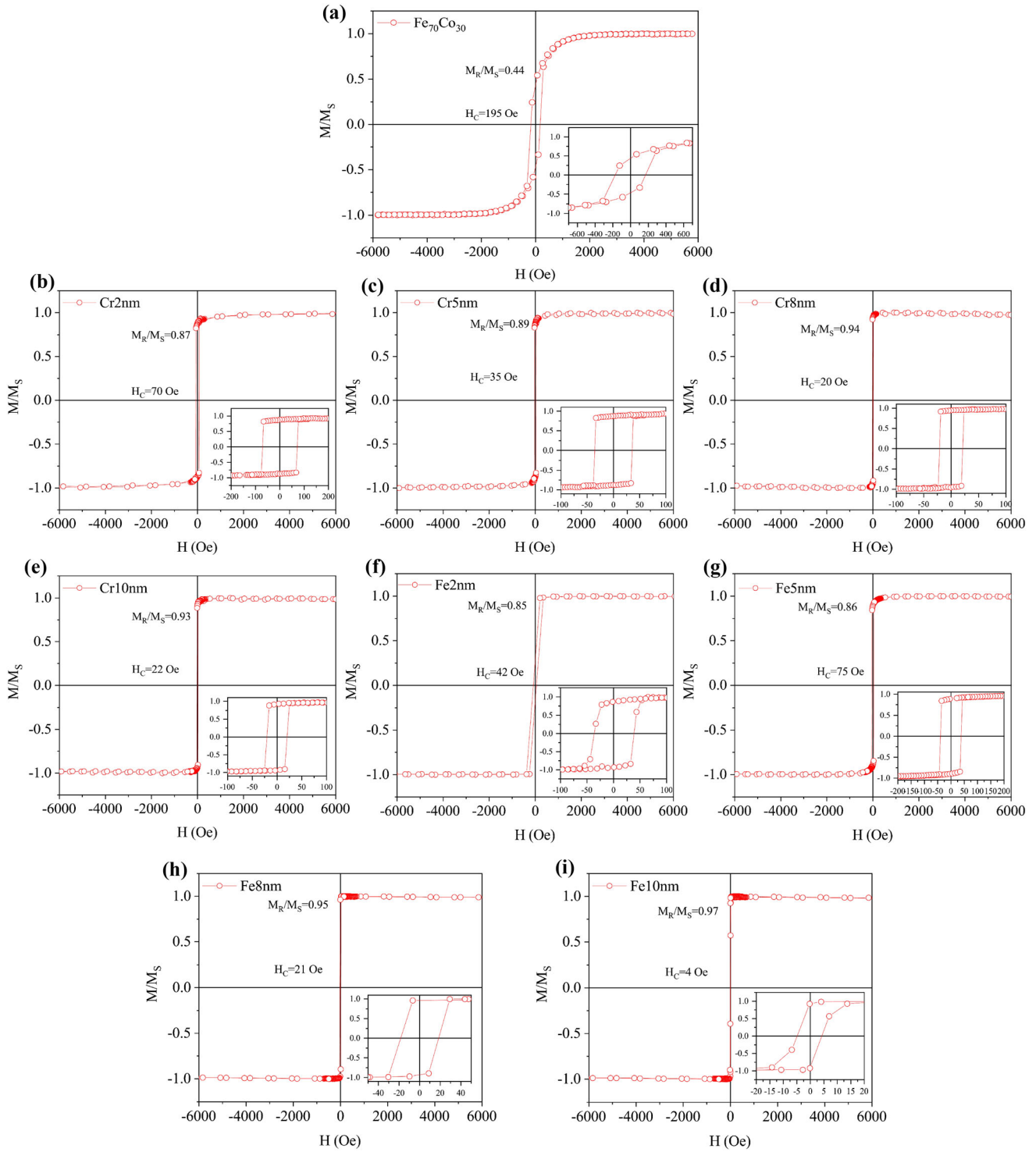


Fig. 3. Magnetic field dependence of thin films: (a) FeCo, (b) FeCo/Cr (2 nm), (c) FeCo/Cr (5 nm), (d) FeCo/Cr (8 nm) (e) FeCo/Cr (10 nm), (f) FeCo/Fe (2 nm), (g) FeCo/Fe (5 nm), (h) FeCo/Fe (8 nm), and (i) FeCo/Fe (10 nm).

differences, consistent with the random anisotropy model (RAM).⁶¹ The RAM is strictly valid when the magnetic exchange length ($l_{\text{ex}} = \sqrt{\frac{2A}{\mu_0 M_S}}$, A , exchange stiffness constant, and μ_0 , permeability of free

space) is larger than the grain size (or when $D \ll l_{\text{ex}}$), which produces complete averaging of local anisotropy. For FeCo, l_{ex} is on the order of 3-5 nm, while the coherent-domain sizes extracted from D-S/W-H analysis (13–21 nm) exceed this

limit. Thus, the system is not in the full Herzer regime. A more appropriate interpretation is a partial random-anisotropy regime where exchange coupling still averages some local anisotropy fluctuations, but strain and texture effects remain significant. This interpretation explains why the H_C decreases with buffer thickness but does not reach the extremely low values reported in ultrafine-grained FeCo ($D \approx 8$ nm). Our data therefore support a hybrid mechanism: partial RAM behavior combined with buffer-dependent strain relaxation and morphology-driven pinning reduction.

To contextualize the performance of the present FeCo/Cr and FeCo/Fe buffer series, Table II summarizes key parameters from representative literature reports on FeCo thin films with various underlayers, in combinations such as Cu or NiFe. The comparison highlights that soft magnetic properties such as low H_C and refined grain sizes have been achieved previously using Co and other seed layers^{42,47} and that our Fe-buffered films achieve very low in-plane H_C comparable to or better than that in some prior works, while also providing systematically high remanence ratios ($M_R/M_S \approx 0.95$). Meanwhile, non-magnetic buffers such as Pd or Pt also reduce H_C relative to no buffer,⁴⁶ but do not achieve the same M_R/M_S . These comparisons position the current work's combination of a systematic study of buffer thickness and competitive performance within the broader FeCo thin-film literature.

Figure 4a and b reveals the M_R/M_S ratio and H_C values of thin films for various Co/Fe buffer-layer thicknesses, respectively. With the results already showing an increasing trend in the lattice parameter for the Fe buffer layer and a decreasing trend in the lattice parameter for the Cr buffer layer, i.e., increasing or decreasing as the thickness of the buffer layer increases, it was observed that the M_R/M_S ratio and H_C measurements followed the changing buffer layer, predicting that this change would be dramatic in the magnetic measurements when looking at the I_{211}/I_{110} ratio. The development of crystallographic orientation is due to the reduction

of surface and strain energies.^{27,28} Grain-size and exchange coupling the Fe-buffered series shows more uniform, smaller grains and a stable grain-size distribution, consistent with robust inter-grain exchange averaging (RAM), which yields a reproducible M_R/M_S . Concerning microstrain and domain-wall mobility, the W-H microstrain

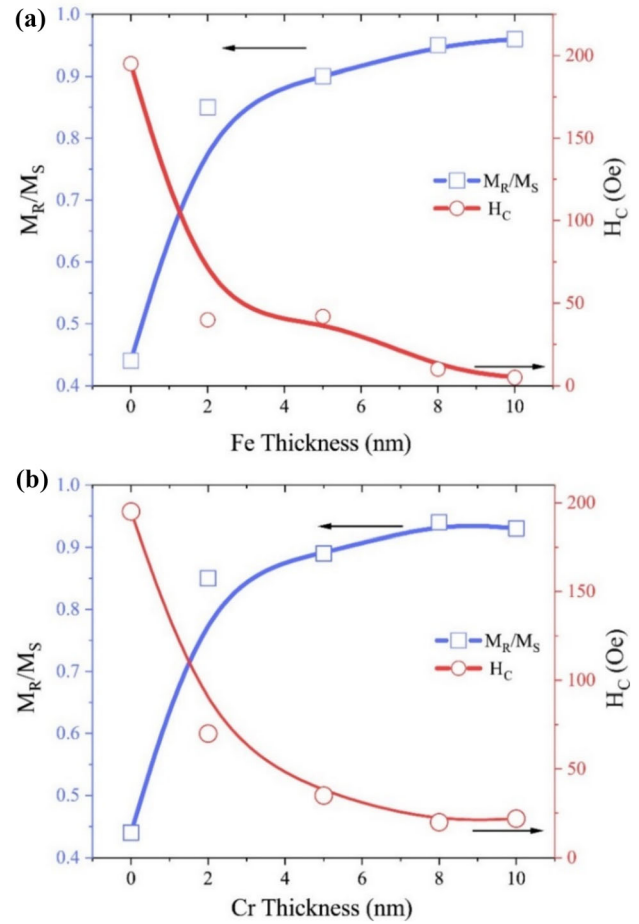


Fig. 4. M_R/M_S ratio and H_C values of thin films within various buffer-layer thicknesses for (a) Fe and (b) Cr layers.

Table II. Comparison of FeCo thin films with buffer/underlayers

System	Buffer/underlayer	Thickness	Grain size (nm)	Hc (Oe)	M_R/M_S	References (DOI)
FeCo (this work)	Cr	2–10 nm	< 18	25-195	0.92	This work
FeCo (this work)	Fe	2-10 nm	< 15	4-200	0.95	This work
Fe ₇₀ Co ₃₀	Pd	~ 1 nm	Refined	~ 15	-	46
Fe ₇₀ Co ₃₀	Pt	~ 1 nm	Refined	~ 15	-	46
Fe ₆₅ Co ₃₅	Co	2-30 nm	8-74	3-10	-	42
Fe ₆₅ Co ₃₅	Co	5 nm	< 10	3-10	-	47
Fe ₆₅ Co ₃₅	Co	5 nm	< 15 nm	1 Oe	0.98	27
Fe ₆₅ Co ₃₅	Cu/NiFe/Ru	~ 2.5 nm	9-50	7-12	-	45
Fe ₆₅ Co ₃₅	None	-	15-50	27-30	-	40
Fe ₆₀ Co ₄₀	Co	2.5 nm	10	11-32	-	47
FeCoN _x	None	-	Amorphous	7.8	-	44
FeCo	None	50 nm	None	3-160 Oe	-	43

increases modestly with buffer thickness, modifying local anisotropy fluctuations and reducing pinning heterogeneity in the Cr series (Table I). Together, these effects can raise the remanence without substantially increasing H_C when grains act cooperatively (strong exchange coupling) and when a growing textured anisotropy aligns the magnetization in-plane. Thus, in the Cr series, the combined increase of texture and moderated pinning leads to the observed systematic rise in M_R/M_S , whereas in the Fe series, a stable microstructure and consistent exchange averaging maintain a near-constant M_R/M_S . These interpretations are consistent with the random-anisotropy arguments of Herzer and with previously reported buffer-layer effects in FeCo films.

The Fe and Cr buffers modify FeCo growth via distinct interfacial pathways. Fe, being ferromagnetic and crystallographically compatible, promotes laterally coherent nucleation and rapid coalescence, resulting in spherical grains and high exchange continuity across the interface. Cr, by contrast, is non-ferromagnetic at room temperature and possesses a higher surface energy and a distinct BCC lattice parameter; these factors encourage columnar/nanorod-type nucleation and more pronounced in-plane strain relaxation. These microstructural differences drive divergent magnetic behaviors. Fe-buffered films exhibit highly coupled, isotropic grains leading to stable remanence but modest changes in H_C with thickness. Conversely, Cr-buffered films develop anisotropic nanorod-like grains that facilitate domain wall propagation and reduce pinning, resulting in a more pronounced decrease in H_C and a systematic increase in squareness (M_R/M_S).

The stronger reduction of H_C observed in FeCo/Cr films relative to FeCo/Fe can be attributed to the combined action of three mechanisms. First, Cr promotes enhanced in-plane strain relaxation, which reduces the effective magnetoelastic contribution to anisotropy as the buffer thickness increases.^{62,63} Second, the non-ferromagnetic nature of Cr restricts direct interfacial exchange continuity, causing FeCo to nucleate in a more energetically favorable configuration with reduced inherited anisotropy. Third, the higher packing density and elongated grain morphology observed in Cr-buffered films reduce domain wall pinning density and enable smoother wall propagation. Together, these effects explain the monotonic increase in M_R/M_S and the more substantial H_C reduction in the Cr-buffered series. While Fe-buffered films also show H_C reduction, the Fe/FeCo interface partially inherits the magnetic texture and exchange anisotropy, thereby moderating the overall change in H_C .

CONCLUSION

The study demonstrates that buffer-layer engineering can modify the microstructure and soft magnetic characteristics of sputtered FeCo thin films without post-annealing. This study's findings demonstrate that Fe and Cr underlayers (2–10 nm) have distinct and repeatable impacts: Fe buffers promote spherical, exchange-coupled grains with consistent remanence ($M_R/M_S \approx 0.43$ – 0.95) and deliver the lowest in-plane H_C down to ≈ 4 Oe at 10 nm, while Cr buffers favor anisotropic (nanorod-like) surface features, a systematic rise in M_R/M_S with thickness, and a different strain/textural evolution that increases M_S . These results show that buffer thickness in the upper single-digit to 10-nm range is a practical control knob for balancing H_C , M_R , and texture, and the choice of a Fe versus non-ferromagnetic-like (Cr) underlayer should be dictated by the target application. Concretely, Fe buffers are preferable when minimizing H_C and maximizing exchange averaging is the priority (high-frequency inductors, write-head soft underlayers, magnetic sensor cores), whereas Cr buffers are better when enhanced in-plane remanence or texture-driven magnetic alignment is desired (spintronic soft underlayers or packed-density films for integrated devices).

SUPPLEMENTARY INFORMATION

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DATA AVAILABILITY

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

CONFLICT OF INTERESTS

On behalf of all authors, the corresponding author states that there is no conflict of interest.

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