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Effect of anisotropy on the critical antiferromagnet thickness in exchange-biased bilayers

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The dependence of exchange bias on antiferromagnet thickness has been measured in FeF_2/Fe and MnF_2/Fe bilayers. The two fluoride systems have identical crystal structures, similar lattice constants, but anisotropy fields that differ by a factor of 20. Hence, by comparing the antiferromagnetic layer thickness dependence of the exchange bias in the two systems we are able to directly establish the effect of the antiferromagnet anisotropy. We find that the critical antiferromagnet thickness for the onset of exchange biasing is an order of magnitude smaller for the more anisotropic fluoride, confirming the often-used assumption that the anisotropy dictates the critical thickness. By measuring the temperature dependence of the exchange bias and the structural morphology of the layers we are able to prove that the effects we observe are not due to the blocking-temperature thickness dependence or the onset of discontinuity in thin antiferromagnet layers.

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INTRODUCTION

The exchange bias effect (a shift along the field axis of the magnetization hysteresis loop) that occurs at the interface between antiferromagnetic (AF) and ferromagnetic (F) systems, continues to be of interest to condensed matter physicists and materials scientists.[1–4] Despite an intensive recent research effort[2–4] and the impetus provided by magnetic recording applications,[5] we still lack a universal quantitative understanding of the phenomenon.[4] However, great advances have been made in recent years. For example, recent experimental investigations have made the first steps towards probing the spin structure at AF/F interfaces[6–8] as well as providing strong evidence for the formation of domains perpendicular to the interface in the AF layer.[9] On the theoretical side we have new models that realistically take into account the structure and disorder of the materials,[10–14] theories which elucidate more clearly the expected interfacial spin structures,[15,16] and even first-principle calculations of the interlayer exchange interactions.[17] One of the most important aspects of these recent theoretical studies is that they stress the possibility of there being more than one mechanism producing exchange anisotropy, as well as the fact that different models may be more applicable to certain materials systems. In order to tackle some of the open issues in this field we have focused on the model system MF_2/Fe (where M is a transition metal) as the fluorides represent some of the best-studied AF systems, with simple crystal and spin structures, as well as controllable growth in thin-film form.

One aspect of the phenomenology associated with exchange bias, which has been investigated on many occasions, is the influence of the AF layer thickness on the exchange bias and coercivity. This particular aspect of exchange biasing has been investigated experimentally in a variety of metallic AF systems[18–24] as well as oxide systems[23,25,26] and has been the subject of several theoretical investigations.[27–29] Although the exact behavior is often specific to a given system (no doubt this is due in part to the complicating factors of AF microstructure) the central feature which is common to all investigations is that there exists a critical AF thickness t_{AF}^{C}, below which the exchange bias disappears. This behavior can be rationalized within a simple model following the initial interpretation of exchange bias by Meiklejohn and Bean.[1] In essence, exchange bias can only be supported when the anisotropy energy in the AF layer is sufficiently large. Otherwise the magnetization reversal of the F layer simply induces a reorientation of the AF layer surface spins. Within the simple Meiklejohn-Bean model this condition can be written $K_{\mathrm{AF}}t_{\mathrm{AF}} > J_{\mathrm{AF/F}}S_{\mathrm{AF}}S_{\mathrm{F}}$, where K_{AF} is the anisotropy constant of the AF layer, t_{AF} is the thickness of the AF layer, $J_{\mathrm{AF/F}}$ is the exchange interaction across the interface between the AF and F layers, S_{AF} is the spin of the AF interface atoms, and S_{F} is the spin in the interfacial F layer. This results in a critical thickness given by

$$t_{\mathrm{AF}}^{C} = \frac{J_{\mathrm{AF/F}}S_{\mathrm{AF}}S_{\mathrm{F}}}{K_{\mathrm{AF}}}. \tag{1}$$

More recently, Binek *et al.*[27] have generalized the Meiklejohn-Bean approach to obtain an analytical expression for the dependence of the exchange bias on the AF layer thickness,

$$-H_E = \frac{J_{\mathrm{AF/F}}S_{\mathrm{AF}}S_{\mathrm{F}}}{M_{\mathrm{F}}t_{\mathrm{F}}} - \frac{J_{\mathrm{AF/F}}^3 S_{\mathrm{AF}}^3 S_{\mathrm{F}}^3}{8K_{\mathrm{AF}}^2 M_{\mathrm{F}}t_{\mathrm{F}}t_{\mathrm{AF}}^2}, \tag{2}$$

where M_{F} is the magnetization of the ferromagnet and t_{F} is the thickness of the ferromagnetic layer. Solving this equation for $H_E = 0$, i.e., the point where the exchange bias vanishes, gives a critical thickness

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$$t_{\mathrm{AF}}^{C}=\frac{J_{\mathrm{AF/F}}S_{\mathrm{AF}}S_{\mathrm{F}}}{2\sqrt{2}K_{\mathrm{AF}}}, \tag{3}$$

similar to Eq. (1). To summarize, the existence of a critical AF thickness can be easily explained within the simplest models of exchange bias, where the key parameter is the AF layer anisotropy.

The goal of the experiment reported here is to determine whether the simple picture of the critical thickness being determined by the AF anisotropy is actually valid. Ideally we would wish to perform an experiment [to measure $H_E(t_{\mathrm{AF}})$] where the anisotropy of the AF layer can be varied without any change in the crystal structure, spin structure, or the microstructure of the AF layers. An opportunity to do just this is provided to us by the use of the antiferromagnets MnF_2 and FeF_2, which have identical crystal and spin structures, lattice parameters which differ by less than 4%, very similar microstructures in thin-film form[30,31] but, crucially, very different anisotropies.[32] The anisotropy fields[33] of FeF_2 and MnF_2 are 149 and 7 kOe, respectively, i.e., they differ by a factor of 20. In this paper we present the results of experiments where we investigate the AF layer thickness dependence of the exchange bias in both of these systems. After analyzing the thickness dependence of the blocking temperature[34] and the morphology of ultrathin AF layers, both of which can seriously complicate the interpretation of the data, we arrive at the conclusion that the critical AF thickness is indeed controlled primarily by the AF anisotropy. It is important to note that our work, unlike previous investigations, measures the critical thickness *as a function of the AF anisotropy*. It is only by such measurements that we are able to determine if the critical thickness is primarily determined by the AF anisotropy.

EXPERIMENTAL CONSIDERATIONS

The samples were deposited in a high-vacuum chamber by electron-beam evaporation. The base pressure of the chamber was in the low-10^{-8} Torr range, while during the deposition of the fluorides, the pressure was in the high-10^{-7} Torr range. All the layers, in both samples, were deposited at a rate of 1 Å s^{-1} onto MgO substrates except for MnF_2, which was deposited at a rate of 2 Å s^{-1}. These substrates were chemically cleaned, then annealed in the vacuum chamber at 500 °C for 1 h immediately prior to deposition. F layer thicknesses were kept constant at 120 Å, while the AF layer thicknesses were wedged to vary the AF thickness in a single sample. This eliminates sample-to-sample variations. The thicknesses and deposition temperatures (in parentheses) for the FeF_2 sample are 0–300 Å (200 °C) for FeF_2, 120 Å (150 °C) for Fe, and 50 Å (150 °C) for Al. Similarly for the MnF_2 sample they are 250 Å (200 °C) for ZnF_2, 0–700 Å (325 °C) for MnF_2, 120 Å (150 °C) for Fe, and 50 Å (150 °C) for Al. The Al layer prevents the oxidation of the films, while the ZnF_2 is a buffer layer to relax the large lattice mismatch (8%) between MgO and MnF_2, which greatly improves the epitaxy of the sample. An automated sliding shutter positioned very close to the substrate, driven by a vacuum stepper motor with 60 μm precision, is used to produce the MF_2 wedges.

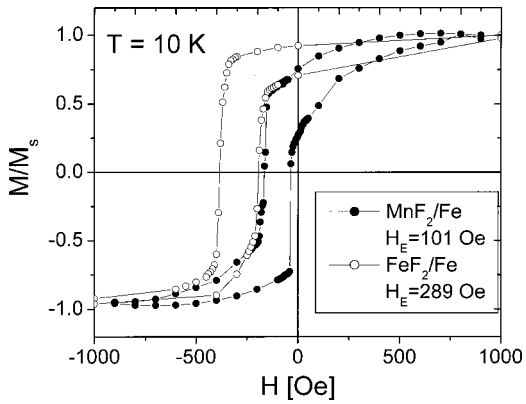

FIG. 1. $T=10$ K hysteresis loops for FeF_2/Fe and MnF_2/Fe, shown as open and solid symbols respectively. The data are plotted as M/M_S, where M_S is the saturation magnetization. The thickness of the AF layer for FeF_2 is 332 Å and 326 Å for MnF_2/Fe. The cooling field $H_{FC}=2$ kOe.

Reflection high energy electron diffraction (RHEED), high-angle x-ray diffraction (HAXRD), grazing-incidence x-ray reflectivity (GIXR), in-plane x-ray diffraction, and atomic force microscopy (AFM) are used to characterize the structure of the samples. Detailed structure and characterization results are provided in previous publications.[30,31] The AFM is performed with a Digital Instruments[35] Nanoscope III Multimode at room temperature using contact mode; AFM images and line scans are used to provide root-mean-square roughness. The interfacial roughnesses of MnF_2/Fe and FeF_2/Fe are similar, and will be discussed in detail later. Likewise, the full width at half maximum (FWHM) of the rocking curves of both MF_2/Fe samples are approximately 2°. The MF_2/Fe wedges were cut into 0.5–1.5-mm slices, which resulted in an error of 12 Å in the averaging of the AF film thicknesses. The wedges were then characterized by superconducting quantum interference device (SQUID) magnetometry from 10 to 300 K. To minimize remnant fields, the superconducting magnet was driven normal immediately prior to measurement of hysteresis loops. Note that no training effects are observed in these samples.

RESULTS AND DISCUSSION

Magnetization hysteresis loops for the MnF_2/Fe and FeF_2/Fe systems are shown in Fig. 1. These two loops were taken at the same temperature (10 K), in a region where the temperature dependence of the exchange bias shows a plateau. Moreover, they were taken at similar thicknesses in a regime where the exchange bias as a function of AF layer thickness is saturated, as will be discussed momentarily. Consistent with previous experiments, the H_E values are considerably larger in the FeF_2/Fe system than the MnF_2/Fe; in this case 290 Oe compared to 100 Oe. It is interesting to examine whether the difference in exchange bias for these two systems can be explained solely on the basis of the different AF anisotropies. This can be estimated using the Meiklejohn-Bean model formula for the exchange

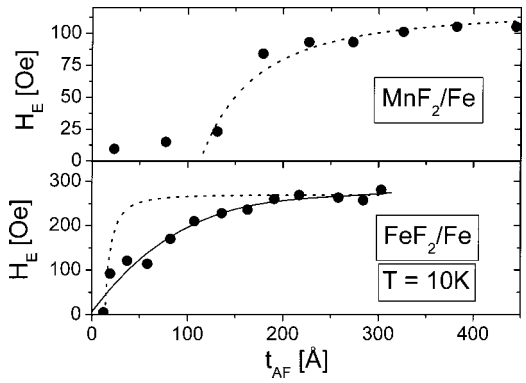

FIG. 2. Thickness dependence of the exchange bias for (a) MnF_2/Fe and (b) FeF_2/Fe, at $T=10$ K. The dotted lines are fits to Eq. (2) (generalized Meiklejohn-Bean model). The solid line is simply a guide for the eye. $H_{FC}=2$ kOe.

bias energy, modified by the factor $(K_{\mathrm{AF}})^{1/2}$ that arises in the Mauri model[36] and the Malozemoff model[37] (see Ref. 4 for a review) due to the introduction of the AF domain wall width, which varies as $(K_{\mathrm{AF}})^{-1/2}$. This leads to $H_E = 2(A_{\mathrm{AF}}K_{\mathrm{AF}})^{1/2}/M_{\mathrm{F}}t_{\mathrm{F}}$ for the Mauri model and $H_E = 2z(A_{\mathrm{AF}}K_{\mathrm{AF}})^{1/2}/\pi^2 M_{\mathrm{F}}t_{\mathrm{F}}$ for the Malozemoff model, where A_{AF} is the spin stiffness in the AF material and z is a constant of order unity. Taking into account the factor of 20 difference in the anisotropy fields for the two fluorides, and the two spin values ($S_{\mathrm{AF}}=\frac{5}{2}$ for MnF_2 and $S_{\mathrm{AF}}=2$ for FeF_2), we would expect that the FeF_2/Fe exchange bias values are about a factor of 4 larger than the MnF_2/Fe values. This compares quite reasonably to the experimental result of a factor of 3, although it is worth noting that different values of $J_{\mathrm{AF/F}}$ could be present[31] as well as the complicating effects of differing AF microstructure. In this case, however, where the growth methods, crystal structures, and the results of extensive structural characterizations are so similar, it is safe to assume that these effects are minimized.

The primary result of this work is shown in Fig. 2, which displays the AF thickness dependence of the exchange bias for both MnF_2/Fe and FeF_2/Fe. Both systems show a monotonic variation of H_E with t_{AF}, and a saturation at large AF layer thicknesses. Moreover, both systems appear to have a critical AF thickness where the bias falls to zero or becomes negligibly small, as expected. For the case of MnF_2/Fe this appears to occur at ~140 Å, while for FeF_2/Fe the exchange bias seems to vanish only at extremely low thickness of the order of 12 Å. Using the simple result of the Meiklejohn-Bean model [Eq. (1) or (3)], which ascribes the value of critical thickness solely to the effects of AF anisotropy, we would expect a variation in critical thickness of ~16, which compares very reasonably to the observed value of ~12. In other words, it would appear that the simple model for the existence of a critical thickness based on the requirement that the anisotropy energy in the AF layer is sufficiently large is actually valid. It is worth noting at this stage that our AF layers do not have a uniaxial anisotropy as assumed in the simple Meiklejohn-Bean model. This is due to the orthogonal twin domains in the fluoride layers. However, the structural situation is the same in both layers and we believe that it is valid to compare the two. Regardless of the microstructure the anisotropy in an individual twin is uniaxial and the final anisotropy resulting from the twinning is still dictated by the size of the ''intrinsic'' uniaxial anisotropy. In any case the data (which assume nothing about the nature of the anisotropy) are quite consistent with the simple model. As an aside we note that the effect of two orthogonal twin domains on perpendicular coupling between the AF and F layers results in a simple fourfold anisotropy as discussed in Refs. 38–40.

We also attempted to fit the full $H_E(t_{\mathrm{AF}})$ behavior using the generalized Meiklejohn-Bean model expression [Eq. (2)]. Examining Eq. (2) we see that the first term in the equation represents the saturation value of the exchange bias at large AF thickness, given by $J_{\mathrm{AF/F}}S_{\mathrm{AF}}S_{\mathrm{F}}/M_{\mathrm{F}}t_{\mathrm{F}}$, i.e., the standard expression for H_E. The second term, which varies as $1/t_{\mathrm{AF}}^2$, is responsible for the reduction in H_E as $t_{\mathrm{AF}}\rightarrow 0$. Given that we know the H_E value as $t_{\mathrm{AF}}\rightarrow\infty$, M_{F} (Ref. 41) and t_{F} (Ref. 42): we are left with only one free parameter—the AF anisotropy. Fits to this expression for the MnF_2/Fe case are shown in Fig. 2(a) as a dotted line. As can be seen from the figure the fit is very reasonable, with an extracted K_{AF} value of 7.3×10^5 erg cm^{-3}, which can be compared to the measured single-crystal bulk value of 4.5×10^6 erg cm^{-3}.[43] Although the discrepancy is quite large it should be noted that the literature value[43] is for bulk single crystals, while our value from fitting is for a twinned thin film. A similar fitting procedure (where we fix the large thickness exchange bias value) applied to the FeF_2/Fe data results in the disappointing fit shown by the dotted line in Fig. 2(b). Clearly the model fails to describe the behavior in the strong anisotropy FeF_2 system. This could be due to a number of factors besides the different anisotropy, including structural effects and the fact that the model is unrealistic in that it does not take into account any thickness dependence of the anisotropy constants. Such a thickness dependence would certainly be more significant at low thickness, meaning that the FeF_2 $H_E(T)$ would be more strongly affected than the MnF_2 $H_E(T)$, as observed.

Despite the apparent excellent agreement between the observed effects of AF anisotropy on critical thickness and the simple model, there are a number of difficulties which arise when interpreting the data of Fig. 2. Firstly, as previously pointed out by Ambrose and Chien[18] the AF thickness dependence of the exchange bias can be strongly influenced by the thickness dependence of the blocking temperature T_B. The blocking temperature is known to decrease with decreasing thickness, which is conventionally interpreted in terms of finite size effects that reduce the intrinsic Néel temperature (T_N) of the AF layer. However, recent work by van der Zaag *et al.*[44] suggests that the decrease in T_B is unrelated to the thickness dependence of T_N, which actually shows *increases* due to a proximity effect with the adjacent F layer. In either case, $H_E(t_{\mathrm{AF}})$ dependencies can be artificially distorted by $T_B(t_{\mathrm{AF}})$. Hence one should determine $H_E(t_{\mathrm{AF}})$ at a constant T/T_B value, or, ideally, the H_E values should be extracted from a low-temperature plateau region, where the exchange bias saturates. Unfortunately, in some systems it is found that the plateau region shrinks with decreasing AF thickness, ul-

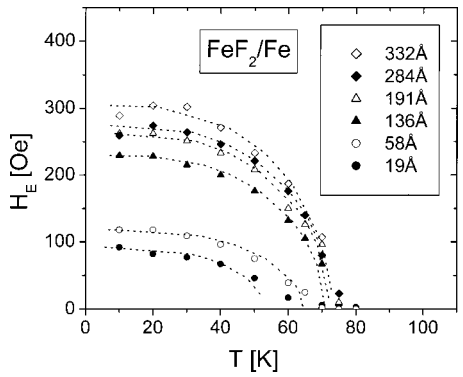

FIG. 3. Temperature dependence of the exchange bias for FeF_2/Fe for various AF film thicknesses (332, 284, 191, 136, 58, and 19 Å). The dotted lines are Brillouin function fits. $H_{FC}=2$ kOe.

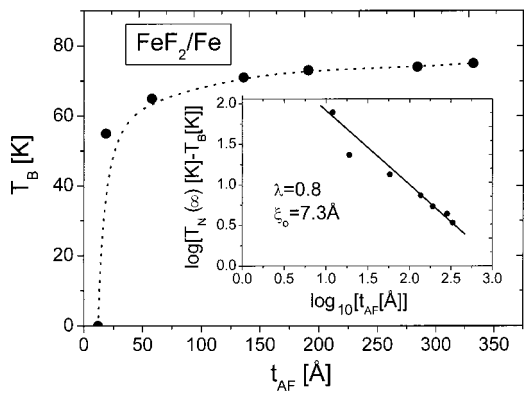

FIG. 4. AF thickness dependence of the blocking temperature for FeF_2/Fe. The dotted line is a guide to the eye. Shown in the inset is $\log_{10}[T_N(\infty)-T_B]$ vs $\log_{10}[t_{\mathrm{AF}}]$, along with the values resulting from a fit to Eq. (4). $T_N(\infty)$ for FeF_2 is 78.4 K (Ref. 18). $H_{FC}=2$ kOe.

timately disappearing at low thicknesses.[18]

The second issue that can complicate the interpretation of the data in Fig. 2 is the structural morphology of the layers at low AF thickness. This problem is particularly acute for our FeF_2/Fe data, where the critical AF thickness is extremely low (~12 Å). It is quite possible that the AF layers are not even continuous at these thicknesses, and that the rapid reduction in exchange bias (below 100 Å) is due to the onset of discontinuity. To tackle these two issues we have determined the effect of thickness on the temperature dependence of the exchange bias (hence extracting the thickness dependence of the blocking temperature) and examined the morphology of ultrathin AF layers with scanning probe microscopy. We will now discuss the results of these investigations in turn.

The temperature dependence of the exchange bias for six samples of FeF_2/Fe with AF layer thicknesses in the range 332–19 Å is shown in Fig. 3. As the thickness is reduced the exchange bias and the blocking temperature both monotonically decrease although the general form of the temperature dependence is unchanged. In particular, the plateau region where the H_E saturates at low T ($0<T<30$ K) is preserved for all thicknesses, rather than crossing over to a more linear dependence of $H_E(T)$ at low AF thickness.[18] In addition to this, the actual reduction in T_B with t_{AF} is very weak, with the blocking temperature of a 19-Å film being as high as 55 K. Before analyzing the $T_B(t_{\mathrm{AF}})$ data in detail, it is worth mentioning immediately that the existence of a plateau in $H_E(T)$ down to $t_{\mathrm{AF}}=19$ Å and the very weak effect of thickness on the blocking temperature mean that the reduction in $H_E(T)$ for FeF_2 above 19 Å (see Fig. 2) is not due to the thickness dependence of the blocking temperature. Having said this, the fact that $H_E(t_{\mathrm{AF}})$ vanishes at a t_{AF} value very close to the point where T_B goes to zero, and the fact that this thickness is on the order of a few monolayers of FeF_2, make it extremely difficult to determine the precise reason for the vanishing of H_E.

$T_B(t_{\mathrm{AF}})$ were extracted from the data of Fig. 3, by fitting with a Brillouin function and are shown in Fig. 4. This simple fitting procedure has been successfully applied to MnF_2/Fe previously[31] and seems to indicate that the temperature dependence of the magnetic order parameter in the AF layer can be well described by a Brillouin function. As previously mentioned, it is not clear in exchange-biased systems whether the thickness dependence of the blocking temperature really reflects the finite size effects in the Néel point.[44] *If* $T_B \approx T_N$, as is the case for thick fluoride layers, then T_B should follow a finite size scaling law of the form

$$\frac{T_N(\infty)-T_B(t_{\mathrm{AF}})}{T_N(\infty)}=\left(\frac{\xi_0}{t_{\mathrm{AF}}}\right)^{\lambda}, \tag{4}$$

where $T_N(\infty)$ is the bulk Néel temperature [78.4 K for FeF_2 (Ref. 30)], ξ_0 is the zero-temperature magnetic correlation length, and λ is the so-called "shift exponent."[45] The inset of Fig. 4 shows a plot of $\log_{10}[T_N(t_{\mathrm{AF}}\to\infty)-T_B(t_{\mathrm{AF}})]$ vs $\log_{10}[t_{\mathrm{AF}}]$, which will allow us to assess whether the scaling form actually describes the data and allows a simple extraction of the constants ξ_0 and λ. As can be seen from the inset, the data are well described by this functional form with $\xi_0=7.3$ Å and $\lambda=0.8$. Although the short correlation length would appear consistent with the fact that T_B only approaches zero at very low thickness, the extracted value of the shift exponent seems unphysical. It is well known theoretically that the Ising and Heisenberg predictions for this exponent are 1.56 and 1.42,[45] while previous measurements on AF layers, such as CoO,[45] NiO,[25] $Co_{0.5}Ni_{0.5}O$,[25] FeMn,[46] and IrMn,[20,25] gave experimental values of 1.55, 1.4, 1.65, 1.6, and 1.5, respectively. Moreover, this exponent has been measured in FeF_2/ZnF_2 superlattices, where the layers were deposited in a very similar manner to those in this study.[47] The result was an exponent of 1.61. In summary, although one can fit the thickness dependence of the blocking temperature to a finite size scaling form, the extracted parameters are unphysical. This is further (indirect) evidence that the assumption that $T_B(t_{\mathrm{AF}})$ is given by $T_N(t_{\mathrm{AF}})$ is incorrect, as suggested by van der Zaag *et al.*[44] It should be noted that this in no way influences our conclusion that our observed thickness dependence of H_E is not due to blocking temperature effects. In fact this is clearly demonstrated by the very different forms of $T_B(t_{\mathrm{AF}})$ and $H_E(t_{\mathrm{AF}})$. As a final comment on $T_B(t_{\mathrm{AF}})$ we should point out that this weak thickness

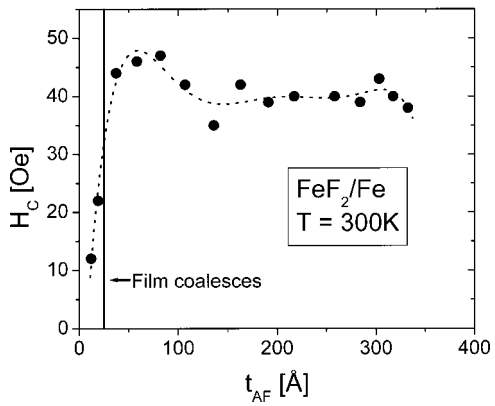

FIG. 5. AF thickness dependence of the coercivity of FeF_2/Fe at $T=300$ K. The vertical solid line is the approximate film thickness where the FeF_2 begins to coalesce. The dotted line is a guide for the eye.

dependence, which results in relatively high blocking temperatures at thicknesses of a few tens of Å, is likely due to the fact that FeF_2 is a highly anisotropic AF material.

The second possible problem that had to be considered was the morphology of the AF layers and the possibility of the onset of discontinuity in ultrathin layers. To investigate this we undertook an AFM study of the surface morphology of MnF_2 and FeF_2 wedges grown in an identical fashion to the ones used for the exchange bias studies. Note that these measurements were made on uncapped layers although we expect the fluoride surfaces to be relatively air stable.[48] Prior to the AFM experiments we were provided with a suspicion of structural changes in ultrathin FeF_2 layers by the magnetometry data at high temperature. Figure 5 shows the AF layer thickness dependence of the 300-K coercivity in FeF_2/Fe, where the effects of the FeF_2 magnetism (maximum $T_N=78.4$ K) do not play a role. The coercivity is roughly independent of the AF thickness down to ~20 Å, where an abrupt decrease occurs. In these samples at 300 K the primary factor that controls the coercivity of the Fe overlayer is the film microstructure. Hence, the existence of a distinct change in behavior around 20 Å is strong evidence that some form of structural change occurs at this AF layer thickness.

The suspicion of the onset of discontinuity at 20-Å FeF_2 thickness is confirmed by the AFM study on FeF_2 wedges, as shown in Fig. 6. This figure summarizes the results of AFM measurements in the interesting regime below 45 Å of FeF_2. Initially, at 10-Å average thickness, the film grows via the formation of small islands of typical lateral dimension 1000 Å, a very common growth mode. This results in a rough surface with typical rms roughness values around 8 Å. As the thickness increases to 20 Å these islands grow in vertical and lateral dimension until the film lies on the verge of coalescence. At this point the rms roughness value reaches a maximum of 18 Å. At higher AF layer thicknesses the film coalesces and begins to form a smooth, continuous film. The roughness decreases, eventually reaching rms values as low as 1.7 Å at 45-Å thickness. The behavior of the rms roughness as a function of average film thickness is shown in Fig.

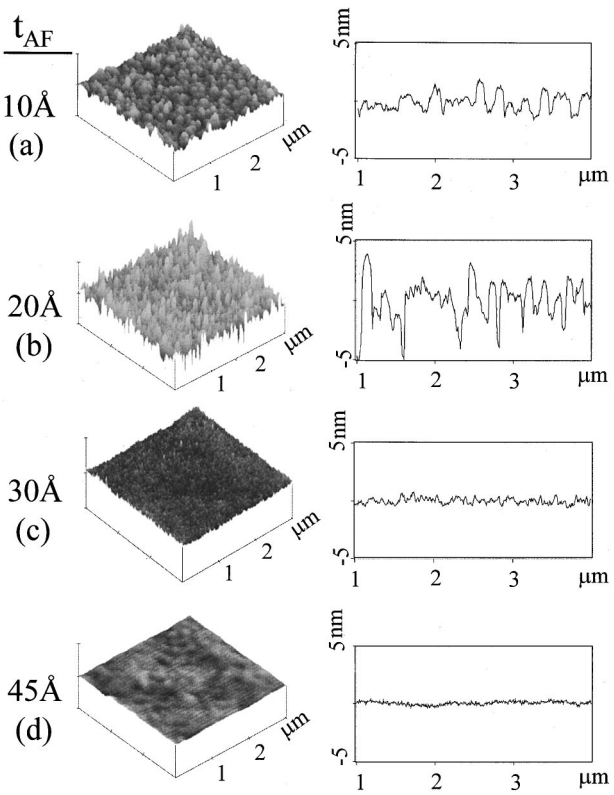

FIG. 6. Three-dimensional AFM images and line scans of the surface of FeF_2. All scans were done over a 3-μm square area of the sample in contact mode. Average AF layer thicknesses of (a) 10 Å, (b) 20 Å, (c) 30 Å, and (d) 45 Å are shown. Coalescence begins at approximately 20–30 Å.

7, where a clear maximum exists near 20-Å thickness (note that the single data point at 20-Å average thickness is the result of averaging over many line scans). It should be noted that MnF_2 samples coalesce at comparable thicknesses to the FeF_2, meaning that the apparent critical thickness of the MnF_2/Fe (i.e., 140 Å) is not at all affected by the structural morphology.

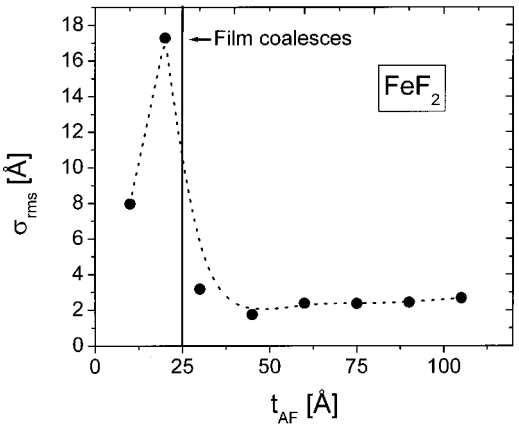

FIG. 7. AF thickness dependence of the root mean square (rms) roughness of FeF_2 obtained from the AFM images shown in Fig. 6. The vertical solid line is the approximate film thickness where the FeF_2 begins to coalesce. The roughness values are averaged over the same 3-μm length scale as the line scans in Fig. 6.

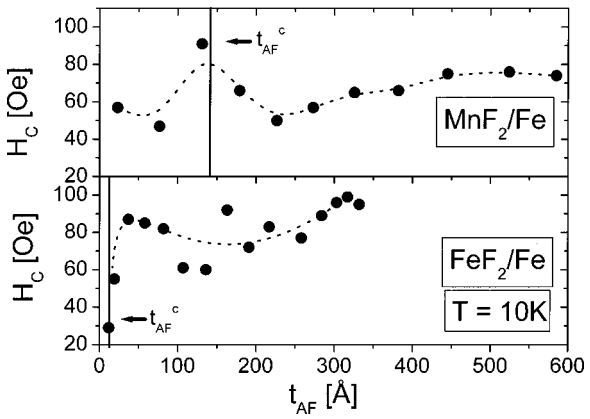

FIG. 8. $T=10$ K AF thickness dependence of the coercivity of (a) MnF_2/Fe and (b) FeF_2/Fe. The t_{AF}^C points are shown as solid lines. The dotted lines are guides for the eye. $H_{FC}=2$ kOe.

The implications of these structural investigations for the data in Fig. 2 are clear. Given that the film is coalesced above ~20 Å, the significant reduction in $H_E(t_{AF})$ that occurs at $t_{AF}>20$ Å is unlikely to be related to structural morphology. Only at $t_{AF}<20$ Å does the structure affect the exchange bias, i.e., masking its *precise* onset. If we were to have a continuous and smooth FeF_2 layer at $t_{AF}<20$ Å, the actual onset of H_E may be even smaller. Therefore, our observed critical thickness should be regarded as an estimate, and the upper limit, for t_{AF}^C in FeF_2. Moreover, the apparent critical thickness of MnF_2 is not affected at all by the onset of discontinuity as these occur at similar thicknesses to the FeF_2 layers, i.e., 20 Å. In summary, the conclusion that the critical thickness is controlled by AF anisotropy remains intact.

It is also worthwhile to consider other possible effects due to the evolution of microstructure with AF thickness. Although our samples are epitaxial, it is quite possible that the twin boundaries play a similar role to grain boundaries in polycrystalline AF materials, i.e., they generate uncompensated spins. However, although the full width at half maximum of the in-plane x-ray diffraction peaks shows a strong sensitivity to deposition temperature, it is not strongly thickness dependent. In other words, the twin domain size displays no significant variation with AF thickness. This is strong evidence that no such effects take place.

As a final point it is worth examining the behavior of the coercivity enhancement, which occurs below T_B and is closely related to the exchange biasing. The AF layer thickness dependence of the 10-K coercivity is shown in Fig. 8 for both MnF_2/Fe and FeF_2/Fe systems. Although the dependence is weak and the data show significant scatter, it seems that in both cases there is a weak reduction in H_C as t_{AF} is reduced, followed by a peak at some thickness value. In the MnF_2/Fe case the peak occurs at ~150 Å, while for the FeF_2/Fe samples it occurs in the region of 45 Å, i.e., in both cases the coercivity reaches a small maximum in the region of the critical thickness where the exchange bias goes to zero. Such behavior has been observed before in several AF/F bilayer systems[2] with AF layers as varied as CrAl,[19,21,22] FeMn,[23,24] NiMn,[23] and NiO.[23,26] As discussed in Ref. 2 an intuitive explanation for the phenomenon is similar to that given for the peak in $H_C(T)$ in the vicinity of the blocking temperature.[2,49] The simple notion is that the vanishing of the exchange bias at low AF thickness is driven by a reduction in the anisotropy of the AF material. As $K_{AF}(t_{AF})\rightarrow 0$, the reversal of the magnetization in the F layer results in a larger effect on the spin structure of the AF layer and, due to their coupling, an increase in the F layer coercivity. This effect eventually disappears at even lower AF thicknesses as the AF order is completely lost—hence the peak structure in $H_C(t_{AF})$. A theoretical modeling of the coercivity in systems with realistic structural disorder by Stiles and McMichael[12] has elaborated on this by pointing out that irreversible changes in the AF spin structure are of importance and that there should exist two regimes of coercive behavior. In one case the losses on sweeping out a hysteresis loop are primarily in the AF layer, while in the other case they are confined to the F layer. The existence of these two regimes was indeed observed in the ferromagnetic layer thickness dependence of the coercivity in MnF_2/Fe, where the extent to which the losses were taking place in the AF layer was probed via the peak in $H_C(T)$.[49] Given the behavior shown in Fig. 8, i.e., a peak in $H_C(t_{AF})$, it is clear that an investigation of $H_C(t_{AF})$ for various ferromagnet layer thicknesses could well shed further light on the two regimes of coercive behavior.

SUMMARY AND CONCLUSIONS

The central result of this work is that we have measured the AF layer thickness dependence of the exchange bias in MnF_2/Fe and FeF_2/Fe bilayers. These antiferromagnets have identical crystal and spin structures, very similar lattice parameters and microstructures, but very different magnetocrystalline anisotropies. Both systems display a critical AF layer thickness, below which the exchange bias vanishes. The critical thickness is an order of magnitude smaller for the more anisotropic fluoride, confirming the simple model where the critical thickness is primarily determined by the AF anisotropy. By measuring the influence of AF layer thickness on the temperature dependence of the exchange bias and the structural morphology of the AF surface we were able to prove that these results are not biased by the thickness dependence of the blocking temperature or the onset of discontinuity in ultrathin films. The AF layer thickness dependence of the coercivity enhancement was also measured and shown to be qualitatively consistent with the simple explanations based on losses in the AF part of the AF/F bilayers.

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