# Antiferromagnetic coupling between layers in Co/Cr multilayer (invited)

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It has been shown by several groups that Fe layers in epitaxial Fe/Cr sandwiches or multilayers (MLFs) are coupled antiferromagnetically (AF) for very small Cr thicknesses. One way this is manifest in the MLF is that for a fixed Fe thickness (30 Å) the hysteresis curves are progressively flatter as the Cr thickness decreases. This is the opposite behavior to that expected if the Fe layers were uncoupled or ferromagnetically coupled. We have observed the same type of behavior in a series of nominally 21-Å Co/$X$ Cr MLFs where $X$ varies from 14 to 57 Å. These MLFs have been well characterized with both x-ray scattering and cross-sectional high-resolution electron microscopy. They are highly polycrystalline, nonideal MLFs with appreciable mixed interfaces. We have determined that Co/Cr MLF have an interface or exchange anisotropy with an average value of $\sim 7$ kOe in a direction opposite to that of the in-plane magnetization. Such an AF interaction is expected from Cr layers having interface thicknesses comparable to that of the pure Cr regions. It provides a natural mechanism for the Cr layers to align the Co layers either AF or FM depending on the local structural details. Both types of alignment are likely to be present in a MLF and the regions having AF alignment will cause the observed flattening of the hysteresis loops. The magnetoresistance of these MLFs has a behavior similar to, but much smaller than, the "giant" magnetoresistance seen in the epitaxial Fe/Cr MLF and single-crystal Fe whiskers. The magnetoresistance in pure Fe has been attributed to the domain structure and it is likely responsible for this effect in the Co/Cr MLF as well.

## I. INTRODUCTION

A number of novel magnetic effects have been seen in multilayer films (MLFs) which are composed of alternating antiferromagnet (AF) and ferromagnetic (FM) transition-metal elements. Some examples are as follows: Ni/Mn MLFs have spin-glass behavior at low temperatures.[1] Co/Cr MLFs have strongly enhanced spin-density waves[2] and a number of well-defined lines at magnetic fields above the intense resonance due to the uniform mode in the FMR spectra.[3] These are believed to arise from exchange interactions of the Co atoms in the "pure" regions with the Cr atoms in or near the sharper interface of the MLF. Various types of magnetic alignment in the layers of MLF composed of rare-earth elements due to long-range RKKY-type coupling between the layers.[4,5] AF alignment of the FM layers in sandwiches or MLF composed of a FM 3$d$ transition element separated by an AF or nonmagnetic transition element. This latter effect is the one we are concerned with in this paper.

This effect has been extensively studied in epitaxially grown Fe/Cr sandwiches and MLF showing AF coupling of Fe layers across the Cr layers.[6–10] In this system it appeared to be necessary to have epitaxial MLF since only FM coupling of the Fe layers was seen in polycrystalline sandwiches.[6] AF ordering has also been reported for epitaxial Co/Cu MLF where the ordering was detected by polarized and unpolarized neutron diffraction measurements.[11] We have also seen the effect in Co/Cr MLFs which are not epitaxial but are highly polycrystalline due to the $d$-spacing mismatch of $\sim 6\%$ for the growth orientations that occur for this system. These Co/Cr MLF have been well characterized by computer fitting of the large-angle x-ray scattering (LAXS) spectra[12] and by cross-sectional high-resolution electron microscopy (HREM).[13,14] Their magnetic behavior has also been extensively studied as described below.

The AF alignment of the FM layers has been attributed to two effects. The first is simply the decrease in magnetostatic energy due to dipolar interactions similar to domain formation. This lowers the total energy of the MLF by aligning the FM layers in an AF configuration. The other effect is the interaction of the layers by exchange coupling either through nearest-neighbors exchange interaction or by longer-range coupling mediated by RKKY-like interactions via the conduction electrons similar to the mechanism causing magnetic alignments in the rare-earth MLF.

At present both the experiments and their interpretation are in a state of evolution, so a brief summary with some comments on the many experiments on the Fe/Cr systems will first be given. We then describe the structural and magnetic behavior of the Co/Cr MLF.

## II. Fe/Cr SANDWICHES OR MULTILAYERS

This system has been extensively studied with a wide variety of experimental techniques. The AF alignment of the Fe layers has been studied in (100)Fe/Cr(100)/Fe(100) sandwiches using light scattering (LS) from spin waves and

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magneto-optic Kerr rotation[6–8] and spin-polarized low-energy electron diffraction (SPLEED).[9] It has also been studied in Fe/Cr ML by magnetization and magnetoresistance measurements.[10]

The Fe layers in the experiments studied by LS[7,8] were each $\sim 100$ Å thick. Here it was observed that an effective FM coupling between the Fe layers separated by a Au interlayer decreases monotonically with increasing Au thickness and disappeared at $d_{\mathrm{Au}} \geqslant 20$ Å. A variation of the Cr layer thickness over the same range revealed that at a Cr interlayer of 4–9 Å the Fe layers had AF coupling. In small external fields in the [100] direction the Fe layers ordered antiparallel with their magnetization perpendicular to the external field. The exchange asymmetry, $A_{\mathrm{ex}} = (I_{\uparrow\uparrow} - I_{\uparrow\downarrow})/(I_{\uparrow\uparrow} + I_{\uparrow\downarrow})$, measured in the SPLEED experiments, was taken as a measure of the remanent magnetization of the surface layers. The results were interpreted as agreeing with the light scattering experiments. These SPLEED experiments provide the only direct information about the interfaces available for this system. Although it is well established that the electronic structure of surface atoms can be quite different from that of interface of bulk atoms, ignoring this and interpreting a decreased value of the exchange asymmetry as indicating the Fe atoms are in an interface, would suggest that the interfaces between the Cr and Fe are 4–5 layers (8–10 Å) thick. This may overestimate the interface thickness somewhat but unfortunately the two usual techniques for measuring interface thicknesses are not applicable to this system. The determination of the interface thicknesses from the height of the satellites in the LAXS spectrum cannot be used because, due to the good lattice matching and same structure of Fe and Cr, the satellites are negligible. Determining the interface thicknesses from HREM images of these MLFs is not feasible, again because they have the same bcc structure and also because the close atomic numbers of Fe and Cr gives little contrast between the two elements. It would be expected that the Fe layers would tend to cluster since the surface free energy of Fe (2.939 J/m$^2$) is considerably larger than that of Cr (2.056 J/m$^2$).[15] This would lead to a propagating roughness in Fe/Cr MLFs with increasingly poor alignment of the layers farther away from the substrate. This could occur in this well-lattice-matched system without appreciably affecting the epitaxial growth. The growth of the films in the SPLEED experiments were monitored *in situ* by measuring the intensities of selected Auger signals from Fe and Cr and looking for kinks in their slopes. However, it has been shown that this technique often does not have sufficient sensitivity to detect clustering.[16]

No AF alignment of the Fe layers was seen in nonepitaxial Fe sandwiches with interlayers of V, Cr, Cu, Ag, Au, or Pd, which were deposited on sapphire substrates.[6] Only FM alignment was observed there and it was attributed to exchange interactions.

In a series of epitaxially grown (001)Fe/(001)Cr MLFs having nominal 30-Å Fe layers it was found that the hysteresis loops at 4.2 K with an applied field in the plane of the films showed progressively tilted behavior as the Cr layer thickness was decreased (see Fig. 1 of Ref. 10). That is, larger applied fields were needed to align the magnetization for MLFs having smaller Cr thicknesses. This indicates that there is antiferromagnetic alignment between the Fe layers; this is opposite to the behavior expected for no coupling or for FM interactions between layers. For these couplings the films with a greater fraction of Fe would be expected to have squarer hysteresis loops. Magnetoresistance measurements at 4.2 K on these MLF had "giant" values. A 30-Å Fe/9-Å Cr MLF had almost a factor of 2 decrease in resistance between zero field and the saturated state at $\simeq 20$ kOe. It should be emphasized that the resistance of these MLFs is very large due to interface scattering; for this particular MLF it was 54 $\mu\Omega$ cm at zero field. In Ref. 10 it was suggested that the magnetoresistant effect arises from the spin-dependent transmission of the conduction electrons through the thin Cr layers. A recent calculation has been presented to explain this giant magnetoresistance which solves the Boltzmann transport equation with spin-dependent scattering at the interfaces.[17] An effect of the observed magnitude at 4.2 K was found for an electron mean free path of $\sim 6000$ Å. However, using a free-electron model for the conduction electrons, the measured resistivity of the MLF corresponds to a mean free path of $\sim 15$ Å in these MLFs, comparable to the layer thicknesses. In general the low-temperature resistivity of a MLF of small bilayer thickness is large, corresponding to mean free paths of the order of the layer thickness. It is usually inversely proportional to the layer thicknesses indicating that the main mode of scattering is from the interfaces. Thus a Boltzmann equation approach is not valid for these MLFs.

The magnetoresistance of pure Fe shows a very similar and even larger effect than that seen in these epitaxial MLFs.[18–21] This behavior was extensively studied in the 1960s and the large magnetoresistance at zero or small fields was attributed to the increased resistance due to domain structure. Although the zero-field resistivities of the pure Fe samples are much smaller (typically $10^{-2}$ $\mu\Omega$ cm as compared to 50–100 $\mu\Omega$ cm), similar effects are also likely to be present in the epitaxial MLF. The domain structure of these small layer thickness MLF is probably very complex and on a much smaller scale than that of the pure Fe, thus giving an increased scattering that contributes to the magnetoresistance effect seen in these MLFs. As for the case of pure Fe, the minimum in the resistivity in the MLF corresponds to the same saturation field needed to remove the domains in the hysteresis loops.

## III. STRUCTURAL AND MAGNETIC BEHAVIOR OF Co/Cr MLF

The structure of the Co/Cr MLF described here has been well characterized by analysis of the LAXS spectra and by HREM studies. These results have been reported in detail elsewhere.[12–14] It was found that these MLFs have far from ideal layering. The similar in-plane areas per atom of [110] Cr and [10.1] Co maximizes the number of bonds at the interfaces, thus causing these two orientations to be favored as the growth directions in Co/Cr MLF. However, the well-known tendency of columnar growth in Co/Cr alloys also occurs in the MLF, leading to the presence of some aligned

[00.2] and [10.0] Co as well as [110] Cr crystallites in Co/Cr MLFs with bilayer thicknesses, $\Lambda \geqslant 22$ Å. This can be seen in Table I, where the volume fractions of the coherent Co/Cr crystallites in each film, as defined and described in Ref. 12, are listed in column 4. The fraction of coherent crystallites is seen to decrease with increasing layer thicknesses of the films. We found from cross-sectional HREM images of the MLF that the layers grow within the columns, which are quite vertical. Many regions of the MLF do not contain columns but where they exist they have a lateral extent of $\sim$100–150 Å in these MLFs. The roughness of the layers increases as the film grows, as can be seen in the typical image of the 21-Å Co/23-Å Cr MLF shown in Fig. 1. Such behavior is probably due to the tendency of the Co to cluster during growth due to its surface free-energy (2.709 J/m$^2$) being larger than that of Cr (2.056 J/m$^2$).[15]

Analysis of the LAXS spectra with a trapezoidal composition profile model[12] determined that the average coherence length, $L_{ML}$, of the MLF crystallites prepared under optimum growth conditions (substrate temperature $\sim$310 K and a deposition rates of 1 Å/s) varied in the manner seen in Table I, where we list the $L_{ML}$ values in column 3. $L_{ML}$ peaked at a bilayer thickness of $\Lambda \sim 30$ Å for an equal layer thickness series. In contrast, for a 21-Å Co/$X$-Cr series, $L_{ML}$ continually increased over the series as $X$ varied from 14 to 57 Å. This occurs because, due to the complex structure of hcp Co, the coherence of the MLF breaks down in the Co layers for thick Co layer thicknesses. If the Co thickness is kept less than $\sim$25 Å, the coherent MLF crystallites can grow to a larger average size in the growth direction.

The interfaces of the Co/Cr MLF system were found to be asymmetrical with the smaller Co atoms preferentially diffusing and/or penetrating into the more open Cr layers during deposition. Thus most of the interface is on the Co depositing onto Cr side in a bilayer of the MLF. The total interface thicknesses, $I_T$, are indicated in Table I where the second column gives the thickness in angstroms of the "pure" Co region, $t_{Co}$/the total interface thickness/the thickness of the "pure" Cr region in each bilayer, $t_{Cr}$. As seen in Table I, for the nominally equal layer thickness MLF the interface thicknesses increased from $\sim$8 Å (4 layers) for $\Lambda = 14$ to $\sim$30 Å (15 layers) at $\Lambda = 60$ Å. It remained at this value for $\Lambda > 60$ Å due to the limited coherence lengths of these MLFs. For the nominally 21-Å Co series essentially all the Co is seen to go into the interfaces for $\Lambda > 50$ Å. Since the interfaces are often seen to be an appreciable fraction of the bilayer thickness, it is necessary to have a knowledge about their shape and extent in order to be able to interpret magnetic measurements of the MLF.

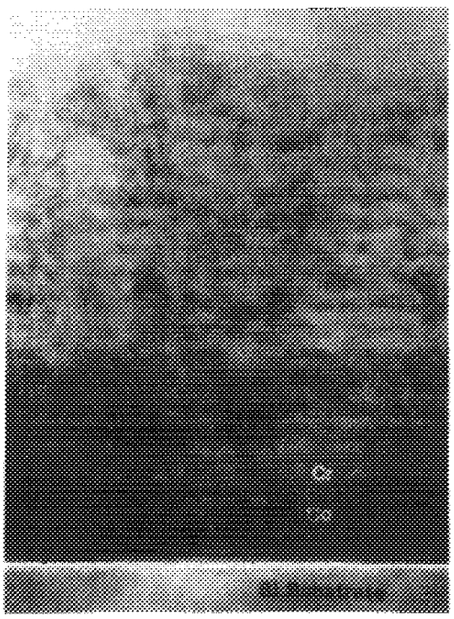

FIG. 1. Cross-sectional high-resolution electron microscope image of the 21-Å Co/23-Å Cr MLF. The darker layers are Co and lighter layers Cr. The Si substrate with its lattice fringes are at the bottom. Note the propagating roughness of the layers and the columnar structure going through the layers.

TABLE I. Structural parameters of Co/Cr MLF. $d_{Co}$ and $d_{Cr}$ indicate the equivalent total thicknesses of the Co and Cr layers for a MLF with sharp interfaces. $t_{Co}$ and $t_{Cr}$ are the thicknesses of the pure Co and Cr regions. $I_T$ is the total interface thickness in a bilayer. $L_{ML}$ is the average length in the direction of growth of the crystalline having a coherent multilayer structure. %MLF indicates the "volume fraction" of the film that is composed of crystallites having a coherent multilayer structure.

| $d_{Co}/d_{Cr}$ (Å) | $t_{Co}/I_T/t_{Cr}$ (Å) | $L_{ML}$ (Å) | %MLF |
|---|---|---|---|
| 6/6 | 2/8/2 | 138 | 100 |
| 6/8 | 2/8/4 | 125 | 100 |
| 12/14 | 2/20/4 | 194 | 99 |
| 16/18 | 6/20/8 | 190 | 98 |
| 19/21 | 8/22/10 | 187 | 97 |
| 22/19 | 11/22/8 | 187 | 97 |
| 28/29 | 16/24/17 | 155 | 93 |
| 34/31 | 19/30/16 | 143 | 85 |
| 48/51 | 34/28/37 | 127 | 77 |
| 100/100 | not a MLF structure | | |
| | Nominal 21-Å Co layers | | |
| 21/14 | 13/16/6 | 159 | 90 |
| 21/23 | 10/22/12 | 165 | 97 |
| 20/30 | 6/28/16 | 192 | 93 |
| 25/43 | 2/46/20 | 191 | 86 |
| 26/50 | 0/52/24 | 210 | 80 |
| 29/57 | 4/50/32 | 210 | 71 |

## A. Hysteresis curves of Co/Cr MLF

### *1. Equal-layer-thickness MLF*

We have measured the hysteresis loops using a SQUID susceptometer at temperatures varying from 4.5 to 300 K with applied fields, $H_{ext}$, up to 25 kOe. We observed that the magnetization lies in the plane of the film for all the equal-layer-thickness MLFs having layer thicknesses greater than 6 Å. The hysteresis curves for perpendicular and parallel applied fields of a 6-Å Co/6-Å Cr MLF were essentially similar, indicating that the magnetization direction in this film had no preference. In general, $M$-$H$ loops of these MLFs with parallel applied field were quite square for MLFs with $\Lambda > 60$ Å. MLFs with $\Lambda < 60$ Å had hysteresis loops that

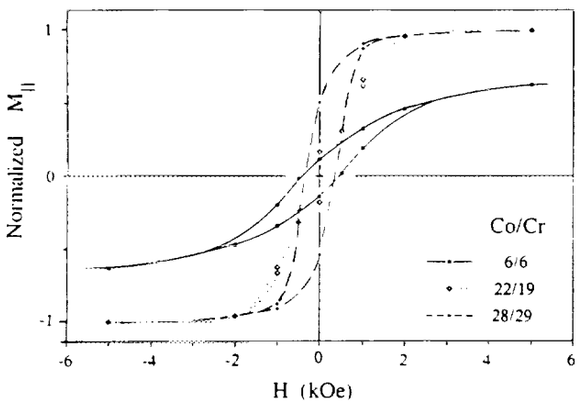


FIG. 2. Parallel applied field hysteresis curves at 4.5 K for several equal-layer-thickness MLFs. For this series the flattening of the loops with decreasing layer thickness is due to a smaller fraction of the Co atoms being in the pure Co regions of the MLF and a larger fraction being in the interfaces as the layer thickness decreases. The Co/Cr thicknesses are listed in Å at the lower right-hand corner.

progressively flattened out as $\Lambda$ decreased. This can be seen in Fig. 2 and Table II where a number of the measured $M$-$H$ loops parameters at 4.5 K are listed. The magnetization saturation values, $M_s$, were taken as the magnetization measured with the applied field, $H_{ext} = 25$ kOe, are listed in column 2 of Table II. These values are the as-measured values without any corrections for the amount of Co in the MLF. The ratios, $m_r$ (in %) of the remanent magnetization to $M_s$ for parallel $H_{ext}$ are listed in column 3. As can be seen, the $M$-$H$ curves tend to become squarer as the Co layer thickness increases. The hysteresis loops of MLFs having $\Lambda < 60$ Å were composed of two distinct regions: a main body at lower applied fields and a longer tail with a smaller slope at increased $H_{ext}$. This dual-region behavior was even more noticeable for the hysteresis loops taken for a perpendicular applied field. This type of hysteresis curve is typical of thin-layer magnetic MLFs and is often seen in the literature. The anisotropy fields for $H_{ext}$ in the film plane are listed in columns 4 and 5 of Table II for the two regions. Where only one value is given there was essentially no long-tail region. The first value denoted as "main" is that of $H_{ext}$ corresponding to the extrapolated intersection of the slope (through the origin) of the main body with $M_s$. The second value is that of $H_{ext}$, corresponding to a magnetization of 0.95 $M_s$. It is seen that all MLFs with $\Lambda < 60$ Å had hysteresis loops with long tails which progressively flattened out with decreasing $\Lambda$. It is suggested that the lower $H_K$ value arises from the Co atoms in the "pure" Co region of the Co layers while the tail region comes from the Co atoms in the interfacial regions. The dual slope behavior is more dominant in MLFs with smaller Co layer thicknesses because for these MLFs the fractional interface region is greater (see Table I). Thus it is clearly evident that Co atoms in the interface are strongly pinned, making it very difficult to change their magnetization direction. Thus for these equal-layer-thickness MLFs an increased flattening of the $M$-$H$ loops with decreasing $\Lambda$ occurs because of the decreasing thickness of the pure Co region and increasing thickness of the interface region as $\Lambda$ decreases.

TABLE II. Measured hysteresis curve parameters for a parallel applied field. $M_s$ is the magnetization measured in an applied field of 25 kOe. The $m_r$ values are the ratios of the remanent magnetization to $M_s$ for an in-plane applied field. $H_K$ main is the value of the applied field at the intersection of the slope of the main body of the hysteresis loop with $M_s$. $H_K$ (95%) is the value of the applied field when the measured magnetization is 95% of $M_s$.

| Co/Cr (Å) | $M_s$ (Oe) (± 10%) | $m_r$(%) | $H_K$(Oe) Main | $H_K$(Oe) (95%) |
|---|---|---|---|---|
| | Nominal equal thickness layers | | | |
| 6/6 | | 13 | 3350 | 21 500 |
| 6/8 | 110 | 23 | 1300 | 7000 |
| 12/14 | 320 | 11 | 2100 | 3200 |
| 22/19 | 510 | 17 | 1080 | 1900 |
| 28/29 | 380 | 51 | 640 | |
| 34/31 | 490 | 72 | 190 | |
| 48/51 | 540 | 84 | 130 | |
| 100/100 | 530 | 86 | 130 | |
| | Nominal 21-Å Co layers | | | |
| 21/14 | 690 | 21 | 1630 | 4000 |
| 21/23 | 470 | 30 | 700 | 1800 |
| 20/30 | 400 | 36 | 810 | 2400 |
| 25/43 | 380 | 73 | 250 | |
| 26/50 | 325 | 76 | 270 | |
| 29/57 | 340 | 80 | 170 | |

### 2. Nominal 21-Å Co/X-Å Cr MLF

The nominal 21-Å Co series shows an effect of a very different nature. Here, as can be seen in Figs. 3 and 4, the hysteresis loops become flatter as the Cr layer thickness decreases. Figures 3 and 4 show the hysteresis loops at 4.5 and 300 K, respectively. The flattening behavior is seen to persist at room temperature and to be even more noticeable due to

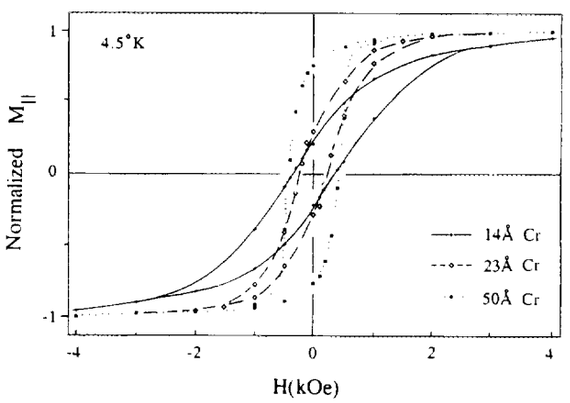


FIG. 3. Parallel applied field hysteresis curves at 4.5 K for three MLFs having a nominal Co layer thickness of 21 Å and varying Cr layer thicknesses. The total Cr layer thickness of each MLF is given in the lower right-hand corner. The flattening of the hysteresis loops with decreasing Cr layer thickness is attributed to the AF alignment between regions of the Co layers in the MLF.

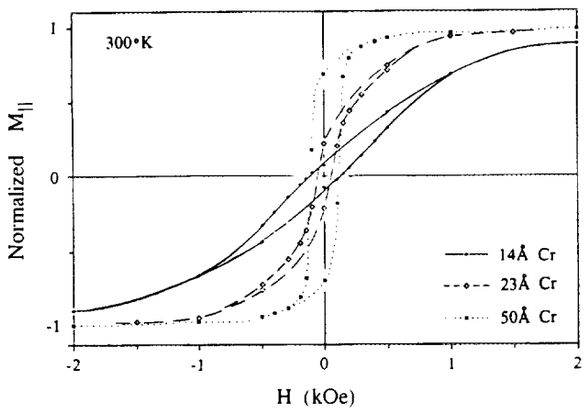


FIG. 4. Parallel applied field hysteresis curves at 300 K for three MLFs having a nominal Co layer thickness of 21 Å and varying Cr layer thicknesses. The flattening of the loops with decreasing Cr layer thickness is due to the AF alignment between regions of the Co layers in the MLF.

the decreased remanent and coercive fields at higher temperatures. For this MLF series, as for the equal-layer series, the thickness of the interfacial region decreases as the Cr layer thickness decreases (see Table I). But in this series $d_{Co}$ is essentially constant so that the fraction of pure Co per bilayer increases as $d_{Cr}$ decreases. Thus for no coupling or FM coupling between the Co layers the hysteresis loops would be expected to become squarer as $d_{Cr}$ decreases. This is opposite to the observed behavior. The observed behavior is of the same type that was seen for the hysteresis loops of the epitaxial Fe/Cr MLF[10] and identified as due to AF coupling between FM layers. We see that such a coupling is also present in these highly polycrystalline, nonideal Co/Cr MLFs having little if any orientational alignment in the plane of the films.

The question of interest is the origin of this coupling. It is well known that due to the large number of itinerant $d$ electrons associated with Cr it tends to align with its moment opposite to the magnetization in dilute alloys with Fe (and probably Co and Ni).[22] As discussed below, there is a large exchange anisotropy of an AF nature acting on the Co atoms in the pure Co regions. This could be responsible for the AF coupling between Co layers. Since the Co/Cr MLFs are far from epitaxial and have rather large interfaces of great structural complexity, it is likely that MLFs containing small Cr layer thicknesses contain regions where the Co layers can be coupled either FM or AF. Probably both types of coupling occur between the Co layers and those that are FM coupled are eventually coupled AF with other regions of the MLF. This AF coupling would show up as seen in the hysteresis loop behavior even if there is also considerable FM coupling between some of the Co layers. The domain structure of these MLFs may also affect the shape of the hysteresis loops since it is expected to be very complex due to the many domain nucleation centers as well as the complex magnetic structure of these MLFs.

### B. Anisotropy fields

The shapes of the $M$-$H$ loops obtained for $H_{ext}$ applied perpendicular to the film were remarkably similar for the whole series of nominally equal-layer-thickness MLFs with layer thicknesses varying over the range of 14–200 Å. Three of these perpendicular hysteresis loops are shown in Fig. 5. The knee of the main loop in these curves occurs at about 4 kOe for all the MLFs. Taking this as the perpendicular anisotropy field, $H_K$, needed to pull the magnetization of the Co atoms in the pure Co regions out of the plane of the film, we can obtain a rough estimate of the interface or exchange anisotropy field, $H_{Ki}$, felt by these Co atoms. We used the following procedure: The saturation magnetization in the "pure" Co regions is obtained by correcting the measured $M_s$ by the ratio of the bilayer thickness to the "pure" Co layer thickness. This latter thickness is taken as the number of layers of the pure Co region plus $\frac{1}{8}$ of the total interface thickness obtained from the LAXS analysis.[23] In this manner we found that all the MLFs with $d_{Co} \geqslant 13$ Å had similar $M_s$ values of $\sim 1400$ Oe, a little less than the $M_s$ value for bulk Co (1475 Oe at 4.2 K). Since the magnetization rotates out of the plane of the film when $H_K + H_{mc} = 4\pi M_s + H_{Ki}$, we can obtain $H_{Ki}$ if we know the magnetocrystalline anisotropy, $H_{mc}$. $H_{mc}$ for [00.2] Co is $\sim 10$ kOe at 0 K, thus it should be $\sim 7$ kOe for [10.1] Co. Using the above values we estimate that $H_{Ki} \sim -7$ kOe. The negative sign means it is oriented opposite to the magnetization. This value should only be taken as a rough estimate since from the LAXS analysis we know that these MLFs contain crystallites of other orientations of Co as well as [110] Cr. The negative sign is in agreement with the known tendency for the Cr atoms to favor AF alignment in alloys with the ferromagnetic transition elements.

### C. Magnetoresistance behavior

In order to more fully investigate the magnetic behavior of the Co/Cr MLF we measured their magnetoresistance. The resistance at 4.5 K was measured using the dc four-

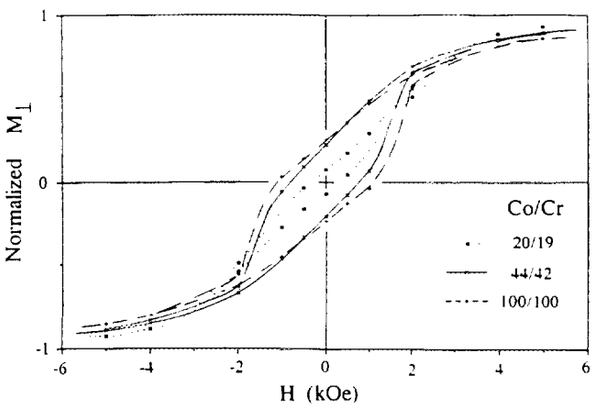


FIG. 5. Typical perpendicular applied field hysteresis curves at 4.5 K for three of the equal-layer-thickness MLF series. The Co/Cr thicknesses in Å are listed in the lower right-hand corner.

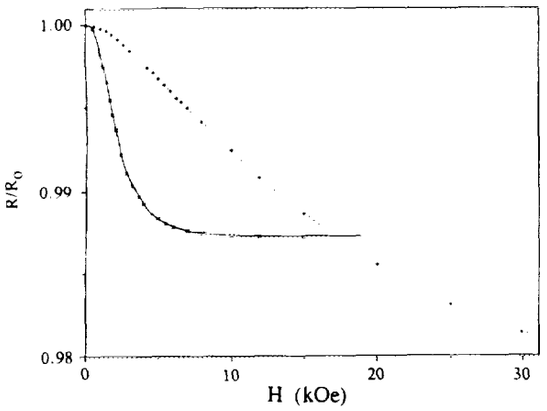


FIG. 6. Magnetoresistance at 4.5 K of a 6-Å Co/6-Å Cr (dotted) and a 21-Å Co/14-Å Cr (solid) MLF. The current and the applied field were in the same direction in the plane of the film. The magnetizations of these MLFs were $\sim 0.95\ M_s$ at $> 22$ kOe and 4 kOe, respectively. The reduction and leveling off of the resistivity occurs as the domains are reduced and removed.

probe method as the field of the SQUID magnetometer was varied. We found a small effect of the same nature as seen in pure Fe or the Fe/Cr MLF. This is shown in Fig. 6 for two Co/Cr MLFs. The magnetoresistance is seen to decrease until the sample is fully saturated corresponding to all the domains being swept out and then it levels off. The behavior with negative applied fields is the same. The observation of a much smaller effect in the Co/Cr MLF than in the pure Fe or epitaxial Fe/Cr MLF is reasonable since the scattering due to domains in these highly nonideal Co/Cr MLF is expected to be small in comparison with that due to the scattering from interfaces, grain boundaries, etc. So the most likely mechanism causing the magnetoresistance in these MLFs would seem to be of a domain nature rather than a spin-dependent scattering effect.

## IV. CONCLUSIONS

It has been seen that the AF alignment of ferromagnetic layers present in epitaxial Fe/Cr MLFs also occurs in polycrystalline Co/Cr MLFs having very complex nonideal structures. These MLFs have been characterized as having considerable interface thickness, multilayer crystallites of 130–200 Å length, and electronic mean free paths comparable to the layer thicknesses. It is also found that the Co/Cr MLFs have an exchange or interface anisotropy of $\sim 7$ kOe in a direction opposite to the magnetization direction. This AF-like anisotropy provides a natural mechanism for producing both FM or AF alignment of adjacent MLF layers depending upon the local fluctuations in the interfacial thickness of the MLF. Such a coupling effect would become more pronounced as the thickness of the Cr layers decreases with the regions having AF coupling, causing the observed flattening of the hysteresis loops.

These MLFs also show a magnetoresistance effect which is much smaller but of the similar nature as the "giant" magnetoresistance seen in pure single-crystal Fe whiskers and epitaxial Fe/Cr MLFs. This decrease in resistance with increasing applied field to a constant value at saturation applied fields has been extensively studied in single-crystal Fe and attributed to the decrease of scattering by the magnetic domains as they are removed by the external field. We believe that this is also the likely mechanism responsible for the magnetoresistance of the Co/Cr MLF.

## ACKNOWLEDGMENTS

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