Magnetoresistance, Micromagnetism and Domain Wall Effects in Epitaxial Fe and Co Structures with Stripe Domains (invited)

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We review our recent magnetotransport and micromagnetic studies of lithographically defined epitaxial thin film structures of bcc Fe and hcp Co with stripe domains. Micromagnetic structure and resistivity anisotropy are shown to be the predominant sources of low field magnetoresistance (MR) in these microstructures, with domain wall (DW) effects smaller but observable (DW-MR \( \lesssim 1\% \)). In Fe, at low temperature, in a regime in which fields have a significant effect on electron trajectories, a novel negative DW contribution to the resistivity is observed. In hcp Co microstructures, temperature dependent transport measurements for current perpendicular and parallel to walls show that any additional resistivity due to DW scattering is very small.

I. INTRODUCTION

The effect of magnetic domain walls (DWs) on the magnetoresistance (MR) of thin films, micro- and nanowires is a topic of great current interest. Recent experimental research has extended early work on 3d transition metal single crystals \( [1] \) to microfabricated structures of transition metals \( [2, 3] \) and transition metal alloys \( [4, 5] \). This topic has also been approached from a number of viewpoints. In nanowires an experimental goal has been to use MR to investigate DW nucleation and dynamics in search of evidence for macroscopic quantum phenomena \( [6, 7] \). While in thin films and microstructures with stripe domains, experiments have focused on understanding the basic mechanisms of DW scattering of conduction electrons \( [2, 3] \). In both cases this experimental work has stimulated new theoretical work in this area, including studies of the effect of DWs on quantum transport in mesoscopic ferromagnets \( [8, 9] \). Independently, a new mechanism of DW scattering was recently proposed which invokes the two channel model of conduction in ferromagnets and spin dependent electron scattering – a starting point for understanding the phenomena of giant MR (GMR) \( [10] \). Another approach has extended the two band model of Ref. \( [11] \) to general band structures and state dependent scattering times, and obtained different results \( [12] \).

In this article we review our recent experimental investigations of patterned epitaxial thin film structures of Fe and Co with controlled stripe domains \( [4, 5] \). Such materials have enabled detailed studies of the physical mechanisms by which the erasure of DWs and micromagnetic structure with applied magnetic fields produces MR. The MR phenomena observed in Fe and Co will be compared and contrasted. Experiments have revealed novel MR effects, including in the case of Fe at low temperature, a reduction in resistivity when domains are present \( [5] \) – in contrast to the increase expected due to DW scattering. This occurs in a low resistivity regime in which both the influence of the internal field on electron trajectories and scattering of electrons at film surfaces are important \( [15] \).

While at low temperature results on Fe and Co are different, at room temperature the MR behavior and the basic physical mechanisms of MR are the same. For example, for fields applied parallel to the easy axis the MR is negative in both Fe and Co microstructures. In initial work on Co thin films with stripe domains this “large” negative MR was interpreted as evidence for the newly proposed DW scattering mechanism, and a giant DW MR \( [3] \). Here MR measurements as a function of the angle of the applied field and magnetic force microscopy (MFM) imaging in conjunction with micromagnetic simulations strongly suggest that this negative MR is mainly a conventional anisotropic MR (AMR) effect – and is thus a bulk scattering effect and not associated with DW scattering \( [6] \). Low temperature measurements on hcp Co microstructures as a function of the angle of the current and walls show that any additional resistivity or MR associated with DW scattering is very small.

II. FABRICATION AND MAGNETIC CHARACTERIZATION

For these studies microfabricated wires were prepared from high quality bcc (110) Fe and hcp (0001) Co epitaxial thin films. These films were produced with an UHV e-beam evaporation system on a-axis (1120) sapphire substrates using seeded epitaxial growth methods \( [4, 5] \). A typical 100 nm thick Fe film prepared in this manner had a residual resistivity ratio (RRR) of 30 and a low temperature resistivity of \( \rho_{0} = 0.2 \mu \Omega \text{cm} \). Co layers of thickness 55 nm, 70 nm, 145 nm and 185 nm (RRR=...
1, \( \rho = 0.16 \mu \Omega \text{cm} \) have been prepared and studied. The films were then patterned using projection optical lithography and ion-milling to produce micron scale bars of 0.5 to 20 \( \mu \text{m} \) linewidth and \( \sim 200 \mu \text{m} \) length for 4 point resistivity measurements.

These films have a strong uniaxial component to the magnetic anisotropy. For (110) Fe thin films, shape anisotropy confines the magnetization to the film plane which contains the easy [001], hard [111] and [110] intermediate magnetocrystalline axes. In contrast, hcp (0001) Co films have a strong uniaxial anisotropy with the magnetic easy axis perpendicular to the film plane.

A competition between the magnetocrystalline, exchange and magnetostatic interactions has been used to produce controlled stripe domain configurations in microfabricated structures. Fe films were patterned into wires with the long wire axis perpendicular to the [001] easy magnetic axis and parallel to the [110] direction, which results in a pattern of regularly spaced stripe domains perpendicular to the long wire axis. Varying the linewidth changes the ratio of the magnetostatic and magnetocrystalline energies to the DW energy and hence the domain size. Minimization of the free energy for such a situation leads to a simple scaling in which the domain width depends on the square root of the wire linewidth \( \mu \). However, experiments reveal metastable domain configurations. Fig. 1 shows magnetic force microscopy (MFM) images of microstructures of systematically varied linewidth in zero field performed at room temperature with a vertically magnetized tip \( [19] \). These images highlight the DWs and magnetic poles at the boundaries of the wires. Images in the left hand column were taken after the wire had been saturated transverse to its long axis, while those in the right hand column were taken after longitudinal magnetic saturation. The domain size is seen to depend both on the linewidth and magnetic history. The latter effect is particularly dramatic in the 2 \( \mu \text{m} \) wire (Fig. 1a and b) where the domain width varies by a factor of 4, from 0.4 \( \mu \text{m} \) after longitudinal saturation to 1.6 \( \mu \text{m} \) after transverse saturation. The domain width can be varied continuously in this range by varying the angle of the in-plane saturating field prior to demagnetization \( [20] \). Note also that the domain width is considerably larger in 20 \( \mu \text{m} \) linewidth wires (\( \sim 6 \mu \text{m} \) Fig. 1e and 1f).

Domain configurations near the sample boundaries have an important influence on the MR, and characterizing their influence is essential to the interpretation of these experiments. Magnetic configurations at boundary surfaces with normal vectors (\( \hat{\mathbf{n}} \)) parallel to the magnetic easy axis (\( \hat{\mathbf{e}} \)) depend on the ratio of the anisotropy to demagnetization energy, \( Q = K/2\pi M_s^2 \). For small \( Q (Q \ll 1) \), flux closure domains (with \( \mathbf{M} \perp \hat{\mathbf{n}} \) to the boundary surface) are favored to reduce the magnetostatic energy, while for large \( Q (Q \gg 1) \) stripe domains which intersect the surface with \( \mathbf{M} \parallel \hat{\mathbf{n}} \parallel \hat{\mathbf{e}} \) are favored to reduce the magnetocrystalline energy density. Both Fe and Co are in the small \( Q \) limit (\( Q_{Fe} = 0.03 \) and \( Q_{Co} = 0.35 \)) and flux closure domains are expected. For the Fe films these occur at the lithographically defined wire edges and for the Co films, which have perpendicular anisotropy, these form at the film top and bottom interfaces \( [17,18] \). Results and micromagnetic simulations for Co in this geometry are discussed below. For Fe microstructures an approximate outline of the domain configuration is sketched in Fig. 1a for a 2 \( \mu \text{m} \) linewidth wire after transverse saturation.

Since current is directed along the Fe wire, there are domains with magnetization \( \mathbf{M} \) oriented both parallel and perpendicular to the current density \( \mathbf{J} \). In order to estimate the MR contributions due to resistivity anisotropy the volume fraction of closure domains (with \( \mathbf{M} \parallel \mathbf{J} \)) has been estimated from MFM images and is labeled \( \gamma \). For the 2 \( \mu \text{m} \) linewidth this fraction is \( \gamma = 0.4 \) after longitudinal saturation and \( \gamma = 0.14 \) after transverse saturation (see Ref. \( [5] \), Fig. 2).

![FIG. 1. MFM images in zero applied field of (a,b) 2 \( \mu \text{m} \), (c,d) 5 \( \mu \text{m} \), and (e,f) 20 \( \mu \text{m} \) linewidth Fe wires. Images in the left hand column were taken after magnetic saturation transverse to the wire’s long axis, while those in the right hand column were taken after longitudinal saturation. The dashed lines in (a) illustrates the flux closure domain configurations observed.](image)

### III. MAGNETOTRANSPORT PROPERTIES

MR measurements were performed in a variable temperature high field cryostat with the applied field oriented in three different orientations: (i) in-plane and perpendicular to the wire long axis (transverse), (ii) in-plane and parallel to the wire long axis (longitudinal), and (iii) perpendicular to the film plane (perpendicular). The sample
A. Fe Microstructures

Fig. 2 shows representative MR results on a 2 \( \mu m \) linewidth Fe wire for in-plane applied fields at both a) high (270 K) and b) low temperature (1.5 K). First consider the MR characteristics at 270 K. For fields transverse to the wire, and thus parallel to the magnetic easy axis, the MR is negative. While in the longitudinal geometry, the low field MR is positive. At fields greater than the magnetic saturation fields (\( H_{s,\parallel} = 0.035 \) T and \( H_{s,\perp} = 0.085 \)T) the resistivity is essentially independent of magnetic field and \( \rho_L(H_s) > \rho_T(H_s) \). At 1.5 K these characteristics change significantly. The transverse MR is now positive, the MR is large and positive above the saturation field and \( \rho_L(H_s) < \rho_T(H_s) \), the resistivity anisotropy is reversed.

This reversal is due to competing sources of resistivity anisotropy. The first is anisotropic MR (AMR) which has its origins in the spin-orbit coupling – the resistivity determined by extrapolation of MR data above saturation to \( B = 0 \) depends on the angle of \( \mathbf{M} \) and \( \mathbf{J} \) and, further, in a crystalline material, may depend on the direction of these vectors with respect to the crystal axes \( \mathrm{2} \). Typically, AMR in transition metals leads to \( \rho_L(B = 0) > \rho_T(B = 0) \), that is domains with \( \mathbf{M} \parallel \mathbf{J} \) have a greater resistivity than those with \( \mathbf{M} \perp \mathbf{J} \). However within magnetic domains, even in the absence of externally applied fields, the \( \mathbf{B} \) field can be large (for Fe \( 4\pi M = 2.2 \) T and Co \( 4\pi M = 1.8 \) T, approximately independent of temperature in the range studied) and the anisotropy of the Lorentz MR can be important. The Lorentz MR depends on the angle of \( \mathbf{J} \) and \( \mathbf{B} \) and is a function of \( B/\rho(B = 0, T) \), the field divided by the zero field resistivity, or equivalently, \( \omega \tau \), the cyclotron frequency times the relaxation time. Since the Lorentz force is proportional to \( \mathbf{J} \times \mathbf{B} \), usually \( \rho_T(B) > \rho_L(B) \).

At low temperature, due to the increase in the relaxation time, \( \tau \), the anisotropy of the Lorentz MR increases. As a result, at a certain temperature, the in-plane \( H = 0 \) resistivity anisotropy changes sign \( \mathrm{2} \). Fig. 3 illustrates schematically this scaling of the resistivity anisotropy with temperature and magnetic field.

Resistivity anisotropy is a conventional source of low field MR. Starting from a multidomain sample, an applied saturating field both erases DWs and reorients the magnetization with respect to the current direction and crystal axes. Since domains with \( \mathbf{M} \perp \mathbf{J} \) and \( \mathbf{M} \parallel \mathbf{J} \) have different resistivities this produces MR. Quite distinct from DW contributions, this low field MR is associated with the electron scattering and orbital effects internal to domains discussed above.

This MR can be estimated within an effective medium model of the resistivity, assuming both that the domain size is greater than characteristic transport lengths (such as the mean free path) and that the resistivity anisotropy is small. The normalized resistivity measured at \( H = 0 \) is given by

\[
R_{\text{meas}}(H = 0) = \gamma R_{L,0} + (1 - \gamma)R_{T,0}
\]

Here \( R_{L,0} \) and \( R_{T,0} \) are the MRs extrapolated from above magnetic saturation to \( H = 0 \) (the dashed lines in Fig. 2),
and $R_{\text{meas}}(H = 0)$ is the measured normalized resistivity (as indicated in Fig. 3). This simple model can account for the high temperature MR. For instance, the negative MR observed in the transverse geometry is due to erasure of higher resistivity closure domains (Fig. 3a). Also, after longitudinal saturation the volume of closure domains is smaller and thus the resistivity at $H = 0$ is lower. Such a model has been employed to analyze the low temperature MR data as well.

![FIG. 4. MR of a 2 μm linewidth Fe wire at $T_{\text{comp}} = 65.5$ K. The extrapolation of the high field MR data to $H = 0$ in transverse (solid line) and longitudinal (dashed line) geometry shows $R_{L,0} = R_{T,0}$, the resistivity compensation at $H = 0$. The resistivity with walls present, $\rho(H = 0)$, is smaller than this extrapolation and indicates that the presence of DWs lowers the wire resistivity. The inset shows this negative DW contribution in the longitudinal geometry as a function of linewidth at $T_{\text{comp}}$.](image)

More directly, due to the competing contributions to the resistivity anisotropy, there is a certain temperature at which the in-plane resistivity anisotropy vanishes. We denote this the compensation temperature $T_{\text{comp}}$ and it is defined as the temperature at which $R_{L,0} = R_{T,0}$. At this temperature the low field MR due to the in-plane reorientation of wire magnetization should vanish. This occurs close to 65 K for the samples investigated. Fig. 4 shows MR results at $T_{\text{comp}}$ for a 2 μm linewidth 100 nm thick Fe wire. The slope of the MR above the saturation field is due to the Lorentz effect and the extrapolation of the high field MR to $H = 0$ (dashed and solid lines) illustrates the resistivity compensation. Fig. 4 shows that there remains a positive low field MR, in both the longitudinal ($\theta = 0$) and transverse ($\theta = 90$) field geometries, and the MR is greatest in the longitudinal geometry in which the DW density at $H = 0$ is largest (Fig. 3b). Since the in-plane resistivity anisotropy is approximately zero at $T_{\text{comp}}$, these results have been taken as evidence for a negative DW contribution to the resistivity [2]. The DW contribution to the MR is calculated at $T_{\text{comp}}$ as $R_d = R_{\text{meas}}(H = 0) - R_{L,T,0}$ and is negative.

By changing the angle of the demagnetizing field the density of DWs has been varied continuously in a single sample between the limiting configurations seen in Fig. 3a and b. The magnitude of the positive MR increases with increasing DW density (Fig. 3). Varying the wire linewidth also varies the density of DWs. The inset of Fig. 4 shows that at $T_{\text{comp}}$ the magnitude of this negative DW contribution to the resistivity decreases with increasing wire linewidth and, hence, decreasing DW density.

![FIG. 5. Cross-sectional view of the magnetic configuration of an Fe wire of thickness $t$, showing the effect of internal fields and surface scattering on the trajectory of charge carriers within stripe domains and DWs.](image)

This effect has been studied further in microstructures of systematically varied film thickness [13]. In films of 200 nm thickness and greater the negative DW contribution to the resistivity is reduced significantly. These results have suggested a novel mechanism by which domains may increase conductivity in thin films. When diffuse electron scattering at the film top and bottom interfaces is important, as in the case of these high quality films at low temperature, the internal field acting on electron trajectories near walls may act to deflect charge from the film interfaces and hence reduce resistivity (Fig. 5). Increasing the film thickness acts to reduce the importance of surface scattering and hence this effect.

**B. Co Microstructures**

We now turn to transport studies of (0001) hcp Co microstructures. Fig. 6 shows MFM images of a 70 nm thick 5 μm linewidth Co wire in zero magnetic field. Images are shown after magnetic saturation: a) perpendicular to the film plane, b) in-plane and transverse to the wire axis, and c) in-plane and along the wire axis. An in-plane applied field can be employed to align DWs in stripes [22]. Fig. 6, b and c shows that DWs can be oriented parallel or perpendicular to the long axis of the wire and thus the applied current, denoted as current-in-wall (CIW) and current-perpendicular-to-wall (CPW) geometries, respectively [13] (as shown in the drawing in Fig. 4). The magnetic structure of a cross-section of the film has been computed with the LLG Micromagnetics
Simulator \[^{23}\]. Details of this simulation can be found in Ref. \[^{6}\]. Calculations for the film thicknesses studied give domain widths which are in excellent agreement with experiment. For the 55 nm thick film a domain width of 66 nm was measured with MFM and the calculated domain width was 64 nm. Fig. 6d shows part of a simulated magnetic cross-section of a 55 nm thick Co element. The arrows indicate the magnetization direction of the stripe and flux closure domains. In-plane magnetized flux closure domains constitute approximately 33 % of the total wire volume. Similarities between the computed magnetic structure of these films in cross-section and MFM images of Fe microstructures in the plane of the film are quite evident (for example, compare this simulation to Fig. 6a).

FIG. 6. MFM images in zero applied field of a of 5 \(\mu\)m linewidth 70 nm thick Co wire after (a) perpendicular, (b) transverse, and (c) longitudinal magnetic saturation. The model shows the orientation of stripe and flux closure domains with respect to the current for (b) CPW and (c) CIW geometries. (d) A calculated magnetic domain cross-section of a 55 nm thick Co element showing out-of-plane magnetized stripe domains and in-plane magnetized flux closure domains.

The general features of the MR of these materials are also similar to those seen in Fe microstructures. Fig. 7 shows MR results for the 3 different field orientations at high (280 K) and low (1.5 K) temperatures. The low field MR is negative for fields applied along the magnetic easy axis (i.e. perpendicular to the film plane), as in the case of Fe at high temperatures, and positive for in-plane applied fields (transverse and longitudinal geometry). At 280 K (Fig. 7a), above the saturation field of \(\sim 1.4\) T, there is a large anisotropy of the resistivity (and a small negative high field MR), with the resistivity largest when the magnetization is in the film plane and parallel to the current \((\mathbf{M} \parallel \mathbf{J})\). As discussed above, this is typical of the resistivity anisotropy due to AMR. Note that the resistivity depends not only on the relative direction of \(\mathbf{M}\) and \(\mathbf{J}\), but also on the direction that \(\mathbf{M}\) makes with respect to the crystal axes, with the resistivity smallest for \(\mathbf{M} \perp \mathbf{J}\) and parallel to the [0001] direction. At low temperature (1.5 K, Fig. 7b) the resistivity is largest above the saturation field in the transverse geometry, with \(\mathbf{M} \perp \mathbf{J}\). As in the case of Fe, and for the reasons already discussed, the in-plane resistivity anisotropy changes sign with temperature.

![FIG. 7. MR data of a 5 \(\mu\)m linewidth 55 nm thick Co wire in the perpendicular, transverse, and longitudinal field geometries at, (a) 270 K and (b) 1.5 K.](image-url)
\[ R_{P,\text{meas}} - R_{P,0} = \gamma \left( \frac{1}{2}(R_{L,0} + R_{T,0}) - R_{P,0} \right) \] (2)

where \( \gamma \) is the volume of in-plane magnetized closure domains. Here \( R_{L,T,P,0} \) are the MR extrapolated from high field to \( H=0 \) (dashed lines in Fig. 7) and \( R_{P,\text{meas}} \) is the normalized resistivity measured at \( H=0 \) in the maze configuration. In this expression, the small volume of in-plane magnetized DW material has been neglected, only the flux closure caps are considered. From the MR measurements shown in Fig. 7a and with \( \gamma = 0.33 \), \( R_{P,\text{meas}} - R_{P,0} \) is estimated to be \( 4.2 \times 10^{-3} \) at \( 280 \) K, in close correspondence with the measured perpendicular MR.

The measured difference between CPW and CIW resistivities (i.e., \( R_{L,\text{meas}} - R_{L,\text{meas}} \)) in Fig. 8 is given in terms of the resistivity anisotropy as

\[ R_{L,\text{meas}} - R_{L,\text{meas}} = \gamma (R_{L,0} - R_{T,0}) \] (3)

which gives \( 1 \times 10^{-3} \) at \( 280 \) K, in close agreement with the experimental value. Although such estimates are qualitative (due to the uncertainties in the material magnetic structure and the applicability of such an effective medium model) they show that the predominante MR effects observed in this material are explicable by film micromagnetic structure and resistivity anisotropy, without the need to invoke DW scattering effects. Thus the simple MR measurements described cannot be used to unambiguously determine the intrinsic effect of DW scattering on resistivity.

In summary, these studies of epitaxial thin film structures have revealed new MR phenomena and elucidated basic mechanism of MR in microstructures with stripe domains. In both Fe and Co, micromagnetic structure and resistivity anisotropy are the predominant sources of low field MR. DWs have a smaller effect on MR. In Fe at low temperatures DWs appear to enhance conductivity via the effect of internal fields on electron trajectories near DWs which act to reduce scattering at film surfaces. In hcp Co, the temperature dependence of the difference between CPW and CIW resistivities is evidence that any effects of DW interface scattering on resistivity are quite small.

These experiments, their analysis and interpretation suggest interesting directions for future research. For example, with the recent theoretical interest in this area, they highlight the necessity of experimentation on materials and nanofabricated structures which will exemplify the transport physics associated with DWs. A further challenge is the extension of such research to nanometer scale epitaxial structures to explore, for example, quantum and ballistic transport phenomena in mesoscopic ferromagnets.
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