INFORMATION TO USERS

This manuscript has been reproduced from the microfilm master. UMI films the text directly from the original or copy submitted. Thus, some thesis and dissertation copies are in typewriter face, while others may be from any type of computer printer.

The quality of this reproduction is dependent upon the quality of the copy submitted. Broken or indistinct print, colored or poor quality illustrations and photographs, print bleedthrough, substandard margins, and improper alignment can adversely affect reproduction.

In the unlikely event that the author did not send UMI a complete manuscript and there are missing pages, these will be noted. Also, if unauthorized copyright material had to be removed, a note will indicate the deletion.

Oversize materials (e.g., maps, drawings, charts) are reproduced by sectioning the original, beginning at the upper left-hand corner and continuing from left to right in equal sections with small overlaps.

Photographs included in the original manuscript have been reproduced xerographically in this copy. Higher quality 6" x 9" black and white photographic prints are available for any photographs or illustrations appearing in this copy for an additional charge. Contact UMI directly to order.

Bell & Howell Information and Learning
300 North Zeeb Road, Ann Arbor, MI 48106-1346 USA
800-521-0600

UMI®
Magnetic and Transport Studies of Thin Film

Ferromagnetic Nanostructures

by

Jun Yu

A dissertation submitted in partial fulfillment of the requirements for the degree of
Doctor of Philosophy
Department of Physics
New York University
September 2000

Professor Andrew Kent, Thesis Advisor
To my family
Acknowledgements

This work would not have been possible without the efforts of many others. First, I would like to thank my advisor, Andrew Kent, for his instruction and support. Andy has not only been an advisor, but also a concerned ear when needed. Thank you Andy for your care.

It has been a delight to work with Ulrich Ruediger. This dissertation could not have come to fruition without his assistance and hard work. I have also enjoyed the friendship with him and his family.

I sincerely appreciate Peter Levy and Shufeng Zhang for many discussions on my experiments. This work has benefited greatly from their insightful theoretical views. I would like to express my gratitude to our collaborators at IBM research labs; Luc Thomas, Stuart Parkin, Robin Farrow and Dieter Weller for providing us thin film samples and micromagnetic simulation.

I thank the staff members at Cornell Nanofabrication Facility, especially Mike Skvarla and David Spencer, for teaching me how to do the micro- and nanolithography. Thanks to Mark Ofitserov for his good lab maintenance. To Yicheng Zhong for the help with the SQUID measurements. To Louisa Bokacheva for helping me operate the cryostat.

And thanks to my wife and my parents, their love has been my foundation.
Abstract

My thesis has involved experimental studies of the magnetic and transport properties of thin film ferromagnetic elements, produced by optical and electron beam lithography. It consists of three related parts: 1) domain wall scattering of conduction electrons, 2) studies of magnetization reversal in thin film nanostructures and 3) exchange biasing in ferromagnetic/antiferromagnetic nanostructures. In each instance, model thin film structures have been developed to study the effects of finite size on these properties, i.e. how the element size, shape and boundaries or interfaces affect their magnetic and transport characteristics.

Electronic transport experiments have been performed on ferromagnetic wires with controlled domain configurations. I have studied three different materials of progressively higher magnetic anisotropy: (110) oriented bcc Fe films, hcp Co films and MBE $L_1_0$ FePt films. By studying the conventional sources of magnetoresistance and isolating the contribution of the domain walls to resistance, I found that domain walls lead to a reduction of resistance in Fe wires, while in Co and FePt wires the resistance is increased with domain walls. I present physical models for these transport characteristics.

I have patterned epitaxial (110) bcc Fe and polycrystalline IrMn/CoFe thin films into elements with rectangular, triangular, and needle-shaped ends. Magnetic force microscope (MFM) imaging and longitudinal Kerr hysteresis loop measure-
ments have been used to investigate the micromagnetic behavior. For Fe elements, the end shape is critical in determining domain nucleation, domain configurations and magnetic hysteresis, which indicates the important role of the demagnetizing field arising from the magnetic charges at the element boundaries. I found that patternning films to micron size does not significantly alter the exchange biasing phenomena in IrMn/CoFe elements. Size effects may become more pronounced at only the deep submicron level, as elements approach the domain sizes (~ 50 nm) in the AF. My results suggest that the random nature of ferromagnetic/antiferromagnetic interactions is essential to model exchange biasing phenomenon.
Table of Contents

Dedication Page .............................................................. iii
Acknowledgements ........................................................... iv
Abstract ............................................................................. v
List of Figures ....................................................................... xi
List of Tables ......................................................................... xv
List of Appendices ............................................................... xvi

Chapter 1 - Introduction ..................................................... 1

1.1 Magnetic interactions ...................................................... 1

1.1.1 Exchange interaction ................................................... 1

1.1.2 Magnetocrystalline energy ............................................. 2

1.1.3 Magnetostatic energy .................................................... 3

1.1.4 Exchange anisotropy .................................................... 4

1.1.5 Zeeman energy ........................................................... 8

1.2 Lateral size effects on magnetic properties ....................... 8

1.2.1 Domain configurations in small ferromagnetic wires ....... 9

1.2.2 Domain wall width ..................................................... 11

1.2.3 Critical size for single domain particles ......................... 11

1.2.4 Stoner-Wohlfarth model ............................................. 12

1.3 Magnetotransport .......................................................... 13
1.3.1 Anisotropy magnetoresistance ........................................... 13
1.3.2 Ordinary magnetoresistance ............................................. 16
1.3.3 Hall effect ........................................................................... 17
1.3.4 Giant magnetoresistance (GMR) ........................................ 18

Chapter 2 - Fabrication ................................................................. 21
2.1 Introduction ............................................................................ 21
2.2 Thin film deposition .............................................................. 23
   2.2.1 Bcc Fe films ................................................................... 23
   2.2.2 Hcp Co films ................................................................. 28
2.3 Making microstructures ......................................................... 31

Chapter 3 Measurement techniques ............................................. 36
3.1 Magneto-optical Kerr effect (MOKE) ..................................... 36
   3.1.1 Macroscopic description ................................................ 36
   3.1.2 Phase modulated MOKE measurement ......................... 43
      3.1.2.1 The phase modulator ............................................... 43
      3.1.2.2 Principles of the technique .................................... 43
      3.1.2.3 Experimental set-up ............................................. 46
3.2 AFM/MFM ......................................................................... 46

Chapter 4 Domain wall scattering ................................................. 53
4.1 Intrinsic domain wall effects ................................................ 53
4.2 Extrinsic domain wall effects ............................................... 57

viii
4.3 Domain wall scattering in Fe films ........................................... 59
  4.3.1 Fabrication and experiments ............................................. 59
  4.3.2 Magnetic characterization .................................................. 60
  4.3.3 Transport properties ....................................................... 70
4.4 Domain wall scattering in hcp Co ........................................... 86
  4.4.1 Magnetic characterization .................................................. 86
  4.4.2 Transport properties ....................................................... 91
4.5 Domain wall scattering in FePt ............................................. 98
  4.5.1 Experiment ................................................................. 98
  4.5.2 Results ................................................................. 99
  4.5.3 Discussion .............................................................. 103

Chapter 5 Magnetization reversal in Fe particles .............................. 108
  5.1 Introduction ......................................................................... 108
  5.2 Fabrication and characterization .......................................... 109
  5.3 Results and discussion ....................................................... 110
    5.3.1 Magnetic easy axis perpendicular to the long particle axis .... 110
    5.3.2 Magnetic easy axis parallel to the long particle axis ............ 127
    5.3.3 Micromagnetic simulations ............................................ 130

Chapter 6 Exchange biasing in polycrystalline thin film microstructures  ........................................................................ 142
  6.1 Models of exchange biasing .................................................. 143
  6.2 Experiment ......................................................................... 144
  6.3 Results .............................................................................. 147
6.4 Discussion ................................................................. 156

Chapter 7 Summary .......................................................... 163

Appendices ................................................................. 166

References ................................................................. 168
List of Figures

Fig 1-1 Magnetic interactions ................................................................. 5
Fig 1-2 Exchange biasing phenomena ...................................................... 7
Fig 1-3 Domain configurations in a ferromagnetic wire ................................. 10
Fig 1-4 Magnetization inside a domain wall ............................................... 10
Fig 1-5 Switching field in the Stoner-Wohlfarth model ................................. 14
Fig 1-6 Domain wall scattering ............................................................... 15
Fig 1-7 Schematic of GMR effect ............................................................ 19
Fig 2-1 Diagram of an e-beam evaporation system ..................................... 24
Fig 2-2 Process for patterning of resist after metal deposition ....................... 25
Fig 2-3 Schematic of the lift-off process ................................................... 26
Fig 2-4 Schematic of the image reversal ................................................... 27
Fig 2-5 Diagram of the Fe surface structure .............................................. 29
Fig 2-6 Hysteresis of a Fe film ............................................................... 30
Fig 2-7 X-ray diffraction of a Co film ..................................................... 32
Fig 2-8 Hysteresis of a Co film ............................................................... 33
Fig 2-9 Optical micrograph of a transport structure ................................... 35
Fig 2-10 SEM of particles ........................................................................ 35
Fig 3-1 The polar Kerr effect geometry ..................................................... 40
Fig 3-2 Elliptically polarized state of the reflected light ................................ 40
Fig 3-3 Other Kerr effect geometries ........................................... 42
Fig 3-4 Schematic of MOKE measurement .................................... 44
Fig 3-5 Experimental MOKE set-up ............................................. 47
Fig 3-6 Basic concept of MFM .................................................. 49
Fig 3-7 MFM image of a Fe wire ............................................... 52
Fig 4-1 Schematic of the similarity between GMR and domain wall .... 54
Fig 4-2 Current distribution in the case of stripe domains ............... 58
Fig 4-3 MFM images of Fe wires ............................................... 61
Fig 4-4 Landau-Lifshitz pattern ............................................... 62
Fig 4-5 Domain width and the volume fraction of closure domains ...... 64
Fig 4-6 Magnetic history dependence of domain configurations of Fe wire ................................. 65
Fig 4-7 Domain width versus applied field angle .......................... 66
Fig 4-8 MFM images of 2 μm Fe wire with external field ............... 68
Fig 4-9 Thickness dependence of domain configurations in Fe wires .......................... 69
Fig 4-10 MR of a 2 μm Fe wire ................................................. 71
Fig 4-11 MR above the saturation field ....................................... 74
Fig 4-12 Kohler plot of a 2 μm Fe wire ....................................... 75
Fig 4-13 MR of 2 μm Fe wire at 65.5 K ....................................... 76
Fig 4-14 MR after annealing ..................................................... 78
Fig 4-15 Angular dependence of MR ........................................ 80
Fig 4-16 Domain wall MR versus temperature ............................ 81
Fig 4-17 MR of 100 and 200 nm thick Fe wires ............................ 83
Fig 4-18 Trajectory of electron inside Fe films ........................................... 85
Fig 4-19 MFM images of Co films ................................................................. 87
Fig 4-20 Magnetic configurations of a Co film after different magnetic histories... 88
Fig 4-21 Domain size versus Co film thickness ............................................. 90
Fig 4-22 MR of 5 μm Co wire ......................................................................... 92
Fig 4-23 Hysteresis of a Co film ................................................................. 93
Fig 4-24 Δν versus temperature ............................................................... 97
Fig 4-25 Optical micrograph of a FePt transport structure ......................... 100
Fig 4-26 MFM images of FePt films ............................................................. 101
Fig 4-27 MR and Hall effect of FePt 1075 ...................................................... 104
Fig 4-28 MR and Hall effect of FePt 1080 ...................................................... 105
Fig 5-1 Hysteresis of 2 μm wide Fe particles ............................................. 113
Fig 5-2 Hysteresis of 0.5 μm wide Fe particles ........................................... 114
Fig 5-3 Magnetization reversal of 2 μm wide particles (Longitudinal) ........... 117
Fig 5-4 Magnetization reversal of a rectangular particle (Transverse) .......... 118
Fig 5-5 Magnetization reversal of a needle-shaped particle (Transverse) ........ 119
Fig 5-6 Domain patterns of 0.5 μm Fe particles ........................................ 122
Fig 5-7 The evolution of the hysteresis loop with the maximum applied field ...... 123
Fig 5-8 Variation of M versus the maximum applied field ............................ 125
Fig 5-9 Nucleation process of 2 μm rectangular Fe particle ......................... 126
Fig 5-10 Hx versus Hz .................................................................................... 129
Fig 5-11 Simulated hysteresis loop (Transverse) ........................................ 133
Fig 5-12 Simulated hysteresis loop (Longitudinal) ........................................... 134
Fig 5-13 Domain patterns of 0.5 \( \mu m \) Fe particles ............................................. 137
Fig 5-14 Magnetic vortices .................................................................................. 138
Fig 5-15 Simulation of defect effects .................................................................. 141
Fig 5-16 Simulation of effects of initial magnetic state ....................................... 141
Fig 6-1 Mauri’s EB model ..................................................................................... 145
Fig 6-2 Spin-flop coupling at FM/AFM interface ................................................ 146
Fig 6-3 Antiferromagnetic domains at FM/AFM interface ................................. 146
Fig 6-4 Hysteresis loops of CoFe and IrMn/CoFe structures .............................. 149
Fig 6-5 Coercivity versus film thickness (unbiased) .......................................... 150
Fig 6-6 Coercivity versus film thickness (EB) ..................................................... 151
Fig 6-7 \( H_c \) and \( H_E \) versus \( 1/t_F \) .................................................................... 152
Fig 6-8 Spin wave frequency versus in-plane angle ......................................... 154
Fig 6-9 Magnetization reversal in a IrMn/CoFe (5 nm) particle .......................... 157
Fig 6-10 Magnetization reversal in a IrMn/CoFe (2 nm) particle ........................ 158
Fig 6-11 Magnetization reversal in a IrMn/CoFe particle (decreasing field) ...... 159
Fig 6-12 MFM image in a IrMn/CoFe (2 nm) film .............................................. 160
List of Tables

Table 4-1 Characteristic data for 2 μm wide Fe wires .................................................. 83
Table 4-2 Characteristic data for 5 μm wide Co wires ...................................................... 97
Table 4-3 Characteristic data for FePt films .................................................................. 100
List of Appendices

Appendix I Lithography process notes ......................................................... 166
Appendix II Image reversal ................................................................. 167
Chapter 1 Introduction

Thin film magnetic nanostructures are a topic of great interest due to their potential applications in high-density data storage and integrated magneto-electronic devices, such as nonvolatile magnetic memory [1-1]. Scientifically, they also provide a great opportunity to study magnetism at small length scales. When element dimensions approach characteristic lengths, such as the exchange length (the length over which the magnetization can change determined by exchange, anisotropy and dipolar interactions), the electron mean free path or spin diffusion lengths, their physical properties are expected to change. In this chapter, basic magnetic interactions and magnetotransport phenomena are reviewed, which provides the background to understand this work.

1.1 Magnetic interactions

1.1.1 Exchange interaction

Magnetic order is a collective effect of spins interacting via the Coulomb interaction and the Pauli Exclusion Principle, i.e. the exchange interaction. In the Heisenberg model, the exchange interaction is

$$-2 \sum_{i<j} J_{i,j} \mathbf{S}_i \cdot \mathbf{S}_j \quad (1-1)$$

where $J_{i,j}$ is the exchange integral, and a positive $J$ means a ferromagnetic coupling that tends to align spins parallel to each other (Fig. 1-1 a). $S_i$ is the total spin of all the electrons bound to the atom, or ion at the lattice site $i$. 1
Though the exchange interaction is very strong, it is a very shorted ranged force between nearest or maybe next-nearest neighbors only. It sets microscopic length scales, such as in the width of a domain wall. The exchange interaction depends on relative angles between spins, not on the spins’ absolute orientation, so is it completely isotropic, and its energy levels do not depend on the direction in space in which the crystal is magnetized.

1.1.2 Magnetocrystalline energy

Real magnetic materials are not isotropic. There are several types of anisotropy, the most common of which is the magnetocrystalline anisotropy, caused by the spin-orbit interaction. The electron orbits are linked to the crystallographic structure, and by their interaction with the spins they make the latter prefer to align along well-defined crystallographic axes. There are therefore directions in space in which it is easier to magnetize a given crystal than in other directions. The difference can be expressed as a direction-dependent energy term. For example, the uniaxial anisotropy energy density can be written as

\[ w_u = -K_1 \cos^2 \theta + K_2 \cos^4 \theta \]  

(1-2)

where angle \( \theta \) is between the anisotropy axis and the magnetization. The coefficients \( K_1 \) and \( K_2 \) are temperature-dependent constants. When \( K_1 > 0 \) and \( K_1 >> K_2 \), it describes an easy axis along \( \theta = 0 \). For cubic anisotropy, the energy density is

\[ w_c = K_1 (\alpha_1^2 \alpha_2^2 + \alpha_2^2 \alpha_3^2 + \alpha_3^2 \alpha_1^2) + K_2 \alpha_1^2 \alpha_2^2 \alpha_3^2 \]  

(1-3)

where \( \alpha \) is the direction cosine between the direction of magnetization and the cube edges.
The magnetocrystalline energy density is usually small compared with the exchange energy. However, the direction of the magnetization is determined only by this anisotropy, because the exchange interaction is indifferent to the direction in space. The magnetocrystalline and exchange energy, together, try to align all the spins parallel to a certain crystallographic direction (Fig. 1-1 b). However, together these two energies are not yet sufficient to explain the subdivision of magnetic materials into domains with different orientations of the magnetization.

1.1.3 Magnetostatic energy

To break crystals into magnetic domains, another energy term needs to be introduced. This term is the magnetostatic energy, which originates from the classical interaction among dipoles,

$$\varepsilon_y = \frac{\vec{m}_i \cdot \vec{m}_j}{|r_{ij}|^3} - \frac{3(\vec{m}_i \cdot \vec{r}_{ij})(\vec{m}_j \cdot \vec{r}_{ij})}{|r_{ij}|^5}$$  \hspace{1cm} (1-4)

It is a self-energy due to the interaction between the magnetic moment of the object and its own field. The magnetostatic potential may be written as

$$U(\vec{r}) = -\int \frac{\nabla' \cdot \vec{M}(\vec{r}')}{|\vec{r} - \vec{r}'|} d\tau' + \int \frac{\vec{n} \cdot \vec{M}(\vec{r})}{|\vec{r} - \vec{r}'|} ds'$$  \hspace{1cm} (1-5)

From which the demagnetizing field can be derived

$$\vec{H}(\vec{r}) = -\nabla U$$  \hspace{1cm} (1-6)

$-\nabla \cdot \vec{M}$ and $\vec{n} \cdot \vec{M}$ can be seen as virtual magnetic charges, which produce the stray field. Then the total magnetostatic energy can be evaluated as
\[ E_M = -\frac{1}{2} \int \mathbf{M} \cdot \mathbf{H} \, dt \] (1-7)

For a uniformly magnetized body, the magnetostatic energy generally depends only on the shape of the object and is anisotropic. This is called shape anisotropy. For example, in the case of a prolate spheroid (Fig. 1-1 c), which is an ellipsoid with two equal axes and one longer axis (defined as the z-axis), energy can be written as

\[ E_M = \frac{V}{2} \left[ N_x (M_x^2 + M_y^2) + N_z M_z^2 \right] = \frac{V}{2} (N_z - N_x) M_z^2 + \text{const} \] (1-8)

This shape anisotropy term has the same mathematical form as the first-order unaxial anisotropy term (Eqn. 1-2), even though their physical origins are different.

To minimize this energy, the magnetic objects will break into magnetic domains. For example, in low-anisotropy particles, flux-closure domain patterns will form (Fig. 1-1 d). In this configuration the normal component of the magnetization is continuous between domains, avoiding "magnetic poles".

1.1.4 Exchange anisotropy

All anisotropy expressions discussed so far are even functions of the magnetization vector, while ferromagnetic (FM)/antiferromagnetic (AF) bilayers often show an equivalent "unidirectional" anisotropy. This lead to an energy of the form:

\[ w_E = -K_{ex} \cos \theta \] (1-9)

where \( \theta \) is the angle between the magnetization direction in the ferromagnet and the preferred direction of the exchange anisotropy.
Fig. 1-1  a) Exchange energy between neighboring spins aligns all spins, b) Magnetocrystalline energy determines the magnetic easy axis, c) Magnetostatic energy tries to align the magnetization along the long axis of an ellipsoid, d) Flux closure domains in low-anisotropy materials.
This anisotropy results from exchange interactions between adjacent FM and AF layers, and gives rise to a series of phenomena, called exchange biasing effects. One phenomenon is a hysteresis loop offset from zero applied magnetic field, which is defined as the exchange biasing field. The second phenomenon usually observed is an increased coercivity of the ferromagnetic layer (Fig. 1-2 a). The exchange biasing effect can be used to control the magnetization in devices, such as spin valves which sense changing magnetic fields through GMR effect. The magnetization of one FM layer is fixed by exchange biasing effect, while the other FM layer rotates freely in an applied magnetic field. Although the exchange biasing is of critical importance to magnetic memories and sensors, mechanisms of exchange biasing are poorly understood.

The simplest model [1-2] assumes that the AF/FM interface occurs at an ideal uncompensated (all spin aligned) plane of the antiferromagnet (Fig. 1-2 b). Both FM and AF layers are treated as single spins, and coupled by an interfacial exchange $J_i$. There is an interfacial energy difference $\Delta \sigma$ per unit area favoring one ferromagnetic orientation over the other. This is just $2J_i/a^2$, assuming a simple cubic structure with lattice size $a$. Then the exchange biasing field is

$$H_E = \frac{\Delta \sigma}{2M_F t_F} = \frac{J_i}{a^2 M_F t_F} \tag{1-10}$$

where $M_F$ and $t_F$ are the magnetization and thickness of the ferromagnet. Unfortunately, reasonable estimates of the relevant parameters give predictions of $H_E$ which are two orders of magnitude too large. For example, considering the system of
Fig. 1-2  a) The drawing shows two common phenomena observed in exchange biased films: hysteresis loop shift and enhancement of the coercivity. The dashed line is the usual hysteresis loop for an unbiased film. b) Spin configurations at ideal uncompensated AF/FM interface
NiFe (permalloy) on antiferromagnetic FeMn, it is plausible that the interfacial exchange interactions should be comparable to the effective exchange interactions of NiFe or FeMn, namely, of order $10^{-14}$ erg. With permalloy thickness 40 nm, magnetization $4\pi M = 10000$ G, and nearest-neighbor distance 2.5 Å, the calculated $H_E$ is 5000 Oe. But the observed hysteresis loop offsets $H_E$ of orders 50 Oe. Further, this model fails to explain biasing when the interface plane of the antiferromagnet is fully compensated (zero net moment).

There are several new models, which can predict the right order of magnitude of the exchange biasing field, and explain the enhancement of the coercivity. These will be discussed in Chapter 6.

1.1.5 Zeeman energy

With an external field the magnetic field energy is simply:

$$E_H = -\int \mathbf{M} \cdot H_\tau \, d\tau$$  \hspace{1cm} (1-11)

1.2 Lateral size effects on magnetic properties

The competition of above energies determines the magnetic properties of the ferromagnet [1-3]. With the miniaturization of magnetic devices, the element boundaries increase the importance of shape anisotropy. Different magnetic behavior will appear when element sizes approach some characteristic lengths associated with magnetic interactions. One topic of my thesis is to explore the finite size effect on magnetic properties, i.e. how the element size, shape, and boundaries or interfaces
affect the magnetic properties. The following are some examples.

1.2.1 Domain configurations in small ferromagnetic wires

Consider a ferromagnetic wire with an in-plane uniaxial magnetocrystalline anisotropy with easy axis perpendicular to the long wire axis (Fig. 1-3). The wire width is \( w \), and its thickness is \( t \). Depending on the ratio of magnetocrystalline to magnetostatic energies, denoted as \( Q = K/2\pi M^2 \), different domain configurations are possible. When \( Q << 1 \), flux closure domains with magnetization parallel to the edges are energetically favored (Fig. 1-3 a). The energy per unit area (i.e. thickness by wire length) is

\[
f(d) = \frac{1}{2} Kd + \gamma \frac{w}{d} \tag{1-12}
\]

where the first term is the magnetocrystalline energy associated with flux closure domains, and the second term is the energy of domain walls. Minimizing the total energy gives the equilibrium condition for domain spacing

\[
d = \sqrt{\frac{2\gamma w}{K}} \tag{1-13}
\]

When \( Q >> 1 \), stripe domains which intersect the boundaries with magnetization perpendicular to the edges are favored (Fig. 1-3 b). The energy per unit area now is

\[
f(d) = \gamma \frac{w}{d} + 1.7M^2d \tag{1-14}
\]

The second term is magnetostatic energy. This gives the domain spacing

\[
d = \sqrt{\frac{\gamma w}{1.7M^2}} \tag{1-15}
\]

9
Fig. 1-3  a) Domain configuration in a ferromagnetic wire with in-plane easy axis perpendicular to the wire when $Q << 1$, b) Domain configuration when $Q >> 1$

Fig. 1-4  Magnetization rotates inside a $180^\circ$ domain wall
From Equation (1-13) and (1-15), it is obvious that by changing the wire width we can control the equilibrium domain size.

1.2.2 Domain wall width

The presence of a boundary between domains involves the rotation of the magnetization vector as illustrated in Fig. 1-4 for a $180^\circ$ wall. To reduce the exchange energy, the magnetization tends to rotate gradually and have large domain wall width. However, the magnetocrystalline anisotropy tends to restrict the width of the wall. With both energies present, the energy of a $180^\circ$ wall per unit area in a uniaxial material is

$$
\gamma = 4\sqrt{A K}
$$

(1-16)

And domain wall width is

$$
\delta = \pi \sqrt{\frac{A}{K}}
$$

(1-17)

in which $A$ is the exchange constant $A = cJS^2/a$ with $c$ the number of atoms per unit cell. For Co with $A = 1.03 \times 10^{-6}$ erg/cm and $K = 4.1 \times 10^6$ erg/cm$^3$, the domain wall width is around 15 nm. Quantity $(A/K)^{1/2}$ is frequently called the “exchange length”, which is the length over which magnetization rotates.

1.2.3 Critical size for single domain particles

As the dimensions of an element are reduced the relative contribution of the domain walls to the energy of the system increases until the point is reached where domain and wall formation is energetically unfavorable; below a certain critical
dimensions, the element behaves as a single-domain particle.

Stoner Wohlfarth [1-4] estimated the critical diameter $D$ for an ellipsoid,

$$D = \left(\frac{2\pi M}{2A/N_s}\right)^{\frac{1}{2}}$$  \hspace{1cm} (1-18)

where $N_s$ is demagnetization factor of the ellipsoid. For Co, value of $D$ for ellipsoids with axial ratio $q = 1$ is 31 nm; when $q = 5$, $D = 76$ nm. For Fe particles, $D$ values are 23 nm for $q = 1$ and 56 nm for $q = 5$.

1.2.4 Stoner-Wohlfarth model

Stoner and Wohlfarth [1-5] proposed a magnetization reversal mode for a single domain particle based on the following premises:

a) Particles are small enough for the exchange energy to hold all spins tightly parallel to each other.

b) The particle will always be oriented such that the total energy $E$ is at a stable or metastable minimum. If, as the external field $H$ is varied, a minimum becomes unstable, then the magnetization orientation switches immediately, going to the nearest minimum.

Consider a particle with uniaxial anisotropy in an external magnetic field, the energy density of this system is

$$E = K \sin^2 \theta - \overrightarrow{M} \cdot \overrightarrow{H}$$

$$= K \sin^2 \theta - M(H_x \sin \theta + H_z \cos \theta)$$ \hspace{1cm} (1-19)
The dependence of $M$ on $H$ is given by the equilibrium condition

$$\frac{\partial E}{\partial \theta} = 0 \quad \frac{\partial^2 E}{\partial \theta^2} > 0$$  \hspace{1cm} (1-20)

It is also deduced that the switching field, at which the metastable state become unstable and the magnetization switches to the nearest minimum, satisfies the follow equation

$$H_z^{2/3} + H_z^{2/3} = \left(\frac{2K}{M}\right)^{2/3}$$  \hspace{1cm} (1-21)

It represents a closed astroid curve of the kind shown in Fig. 1-5.

1.3 Magnetotransport

One topic of my thesis is domain wall scattering on conduction electrons (Fig. 1-6). By applying a large field, one can magnetically saturate the ferromagnet, thus remove the domain walls. Comparing the resistance with and without domain walls, we are trying to determine the contribution to the resistance due to spin dependent scattering of domain walls. It is very important to isolate other sources of magnetoresistance [1-6, 7], such as anisotropy magnetoresistance and Lorentz magnetoresistance, from the contribution of walls.

1.3.1 Anisotropy magnetoresistance

Anisotropy magnetoresistance (AMR) depends on the angle $\alpha$ between the current and the magnetization. For a polycrystalline cubic crystal (fcc, bcc) [1-6], if $\rho_\parallel$ and $\rho_\perp$ are the resistivity values when $\alpha = 0^\circ$ and $\alpha = 90^\circ$, then we define the
Fig. 1-5 Switching field in the Stoner-Wohlfarth model shows a closed astroid curve in the $H_xH_z$-plane.
Fig. 1-6 Electrical current is passing through a multi-domain ferromagnetic wire. Inside each domain wall the magnetization rotates, which may cause spin-dependent scattering of the conduction electrons and change the resistance.
anisotropic magnetoresistivity ratio as
\[
\frac{\Delta \rho}{\rho_0} = \frac{\rho_\perp - \rho_\parallel}{\frac{1}{3} \rho_\perp + \frac{2}{3} \rho_\parallel}
\]  
(1-22)

For arbitrary \( \alpha \), the resistivity \( \rho \) measured in the current direction is predicted to be
\[
\rho = \rho_\parallel - (\rho_\parallel - \rho_\perp) \sin^2 \alpha
\]  
(1-23)

The physical origin of AMR lies in spin-orbit coupling. As a magnetic material is magnetized, the electron cloud about each nucleus deforms slightly and this deformation changes the amount of scattering undergone by the conduction electrons in their passage through the lattice. And it is found that \( \rho_\parallel \) is nearly always greater than \( \rho_\perp \) [1-6, 7]. It should be noted that \( \rho_\parallel \) and \( \rho_\perp \) cannot be measured directly without some form of extrapolation because of the presence of the internal magnetic field.

Due to the presence of AMR, the resistance of a ferromagnet at zero external field depends on the exact domain configurations. So it is history dependent and is not well defined even for a given sample at a given temperature. In the same way, the change of resistivity with field below saturation depends on the magnetization reversal process.

1.3.2 Ordinary magnetoresistance

Ordinary magnetoresistance is associated with the magnetic induction field, which curves the electron trajectories. In a ferromagnet, the induction field \( B \) is
\[
B = H + 4\pi M - H_d
\]  
(1-24)
where $H$ is the external field and $H_d$ is the demagnetizing field. So even without external field, a field $B = 4\pi M$ exists in each domain, which increases the resistivity. In general, the ordinary MR is proportional to $B^2$. In addition, $\rho_\parallel$ is smaller than $\rho_\perp$, which is opposite to AMR.

For the ordinary magnetoresistance, the anisotropy arises from the convoluted nature of the Fermi surface. No magneto resistive effects are predicted using only free electron theory. For example, a transverse field is exactly cancelled by the Hall field, and the electrons experience no additional force in the steady-state [1-6]. To obtain ordinary magnetoresistance, we must go to more complicated Fermi surface than the single spherical surface of the free electron model.

In a ferromagnet above technical saturation the magnetoresistance behavior generally follows Kohler’s rule $[\rho(B) - \rho(0)]/\rho(0) = F(B/\rho(0))$, where $F$ is a function which varies from metal to metal and with the types of scatterer. This rule holds if scattering can be described in terms of a relaxation time $\tau(k)$ and changes in temperature or purity simply alter all $\tau(k)$ by number factor. It is also important that the filled region in $k$ spaces does not change appreciably in size over the range of temperature and purity considered.

### 1.3.3 Hall effect

The various contributions to the Hall effect of ferromagnet can be classified as follows: a) Terms caused by the magnetic Lorentz force and dependent on the field $B$ (ordinary Hall effect); b) Terms dependent on the orientation of the magneti-
zation $M$ (extraordinary Hall effect). The total Hall resistivity is often written as

$$\rho_H = R_0 B + R_s M$$  \hspace{1cm} (1-25)

where $R_0$ is generally a smaller quantity than $R_s$.

Once again, extrapolation of the Hall voltage curve as a function of $B$ from the saturated region back to $B=0$ is necessary to obtain $R_s$. This is easy when $\omega_c \tau << 1$ ($\omega_c$ is the cyclotron frequency and $\tau$ is the electronic relaxation time), but the separation between ordinary and extraordinary effects is difficult when $\omega_c \tau >> 1$ where the ordinary Hall voltage is no longer linear in $B$.

1.3.4 Giant magnetoresistance (GMR)

Reduced dimension will also modify the electronic transport properties. Surface scattering may become important in thin films. In ferromagnetic/non-magnetic metal/ferromagnetic tri-layers, the spin-dependent interface scattering plays an important role in the GMR effect [1-8, 9]. Depending on whether the magnetizations of two ferromagnetic layers are parallel or anti-parallel, the system shows lower or higher resistance (Fig. 1-7). The relative change of resistance can be 10-30%.

In the perpendicular geometry, in which the current flows perpendicular to the film plane, it is a good approximation to assume that the current is carried by two nonintermixing components, spin up and spin down, and that one need only determine the spin-scattering coefficient for each of these components at the interfaces and in the interiors to completely describe the MR behavior of a multilayer
Fig. 1-7  Schematic of conduction in a magnetic multilayer for parallel magnetization in a) and for antiparallel magnetization in b). The upper figures represent electron trajectories in a multilayer for both spin directions. The scattering at the interfaces of the magnetic layers is assumed to be weaker for the majority spin direction. The lower figures represent the equivalent resistor arrays: large and small resistors in parallel in a), with a low resistance path available; large and small resistors in series in each branch (no low resistance path) in b)
structure. The spin relaxation length itself is much longer than the typical 1-10 nm layers in most structures. This means that a given electron can pass through many layers before “forgetting” its spin orientation. Within this length, each magnetic interface can act as a spin filter, and the more scattering interfaces an electron interacts with, the stronger the filtering effect.
Chapter 2 Fabrication

The application of lithographic and other controlled fabrication techniques has given rise to the possibility of exploring magnetism at small length scales. Advanced fabrication techniques can make zero dimensional magnetic elements in which all three dimensions are geometrically constrained. While film thickness can be controlled to monolayer accuracy by thin film deposition techniques, reducing the lateral element size is more challenging. In this chapter I will describe how I fabricated micron and sub-micron ferromagnetic samples. This involved both advanced thin film deposition and lithography methods.

2.1 Introduction

Thin deposited films are important in fabricating microdevices. By depositing a thin film, at least one dimension of the device is relatively straightforward to control with high precision. Single molecular layers (monolayers) a few angstroms thick to a layer several light wavelengths thick (microns) can be formed. There are a number of methods to prepare such films, such as molecular beam epitaxy (MBE), cathodic sputtering, and evaporation in ultrahigh vacuum (UHV) [2-1]. Evaporation in UHV is particularly useful when it is necessary for the film material to remain uncontaminated by residual gases and where the interface between the substrate and the film must have specific properties that could be influenced by contaminants. Pressures of less than $10^{-6}$ Torr are necessary to maintain the substrate
contaminated for a reasonable time after cleaning. Figure 2-1 shows a diagram an electron-beam evaporation system. The material to be evaporated is placed in a recess in a water-cooled copper hearth. An electron current of about 100 - 5000 mA is generated by a tungsten filament, hidden from direct viewing of the evaporant, and accelerated to a high voltage of 3 - 20 KV. The electron beam is magnetically directed onto a small spot on the evaporant material, which melts locally.

After the film deposition, producing nanostructures depends on selective processes that result in materials being positioned in particular areas of a surface [2-2]. This can be accomplished by the removal of material from unwanted areas (etching), or addition of material (deposition). This is done by producing a template in which defined areas of the substrate are protected from or exposed to these processes. These areas make up the pattern. Pattern definition takes place in the resist, which is a thin layer of polymeric material, which is coated on the substrate. The resist is modified by exposure (incident radiation or particles) and development (selective removal in solvent or base), so that it remains in some areas and is removed in others.

To form a metallized pattern on a wafer, there are two methods. The first one is to evaporate or sputter the metal film onto the wafer first, and put a layer of polymer resist on top of it, then pattern the resist. The process flow is showed in Fig. 2-2. After development, an etching process is used to transfer the pattern from resist to the metal layer. A second method is to use a lift-off technique. In lift-off, the resist is patterned first, then the metal is evaporated over the resist. The resist is then dis-
solved away in a solvent, carrying the unwanted metal with it. The lift-off process showed in Fig. 2-3. Between exposure and development, there is an image reversal process, used to create an undercut profile (Fig. 2-4), which is favorable for lift-off.

Depending on exposure tool used, different lateral resolution can be achieved. For projection optical lithography, 1 µm resolution can be achieved. For electron-beam lithography, features below 100 nm can be made.

2.2 Thin film deposition

2.2.1 Bcc Fe films

Epitaxial Fe (110) films have been deposited using UHV e-beam evaporation techniques [2-3]. The substrate, which was single crystal Al₂O₃ [11-20], was cleaned in acetone, isopropyl, and methanol with ultrasonic agitation, then blown dry. The base pressure while heating the substrate immediately prior to deposition was about 10⁻⁸ Torr. A 10-nm Mo seed layer was deposited at a substrate temperature of 900 K. Fe was deposited once the temperature had dropped to 510 K. The deposition rate was around 2 Å/s, which is monitored by a quartz-crystal film thickness monitor. Another 5-nm Mo capping layer was deposited to protect the Fe film.

The thickness of films varies from 25 to 200 nm, and the surface roughness is around 1 nm, measured by atomic force microscope. It has been shown that the (110) Mo planes are parallel to the [11-20] Al₂O₃ surface planes and the Mo [-111] in-plane direction is parallel to the [0001] direction of the sapphire substrate [2-3], which is perpendicular to the wafer flat. Figure 2-5 shows the film surface structure.
Fig. 2-1 Diagram of a 270° magnetized deflective electron-beam evaporation system.
Fig. 2-2 Process flow for patterning of resist after metal deposition.
Fig. 2-3 Process flow for metal lift-off using image reversal
a) before image reversal

b) after image reversal

Fig. 2 – 4 Image reversal forms the undercut profile, which is favorable for lift-off. The metal film is discontinuous over the desired features.
Mo and Fe layers are both bcc, with a lattice mismatch of 9%. Thus, the orientation of the Fe [001] direction is 35 degrees away from the wafer flat. An anisotropic in-plane strain in these films induces an in-plane uniaxial magnetic anisotropy with the easy axis parallel to the [001] direction [2-3]. I have measured the magneto-optical Kerr effect (MOKE) to obtain the magnetic hysteresis loops for these films. Figure 2-6 shows an example measured from a 50-nm thick Fe film. When the external field is applied along the [001] direction, the loop shows a typical square shape of an easy-axis loop. The hysteresis loop with the field along the [1-10] direction has been modeled using the in-plane anisotropy energy;

\[ E = (K_1 + K_u)\sin^2 \theta - \frac{3}{4} K_1 \sin^4 \theta \]  

(2-1)

where \( K_1 \) is the cubic anisotropic constant for bulk bcc Fe, \( K_u \) is the in-plane uniaxial anisotropy constant, and \( \theta \) is the angle between the magnetization and the [001] axis.

Assuming all the spins are parallel and coherently rotate under the external field, we can fit the loop along the [1-10] direction, by minimizing the total energy, \( E = (K_1+K_u)\sin^2 \theta - 3K_1\sin^4 \theta / 4 - MH\sin \theta \), to find the angle at each field. From these fits, \( K_1 \) and \( K_u \) have been determined. In Fig. 2-5, \( K_1 = 6.3 \times 10^5 \) erg/cm\(^3\) and \( K_u = 3.0 \times 10^5 \) erg/cm\(^3\). There is some deviation of this fit from the data near magnetic saturation. This is likely due to small variations in magnetic anisotropy in the film, and particularly the presence of regions of enhanced anisotropy.

2.2.2 Hcp Co films

Using similar techniques, I have grown epitaxial hcp Co films [2-4].
Fig. 2 - 5  This diagram indicates the orientation of the easy axis of magnetization with respect to the sapphire wafer flat.
Fig. 2–6  Longitudinal magnetooptic Kerr effect hysteresis loop measurements on a 50-nm-thick epitaxial (110) Fe film with applied field parallel to the [001] direction (easy axis) and parallel to the [1-10] direction (hard axis). Also included is the fit of the hard axis hysteresis loop.
nm-thick Ru seed layer was deposited on (11-20) sapphire substrate follow by a (0001) Co layer. The Co layer was protected against corrosion by 5-nm-thick Ru capping layer. The films were grown at 680 K. X-ray θ/2θ scans indicate c-axis orientation of the Ru and Co layers. The off-axis x-ray pole figures show that the films have a 6-fold in-plane symmetry and therefore are oriented in-plane with respect to the sapphire substrate (Fig. 2-7). The dominant (0001) hcp structure gives rise to a large perpendicular magnetocrystalline anisotropy (~5×10⁶ erg/cm⁴), which locks the magnetization perpendicular to the layer [2-4]. Figure 2-8 shows the magnetization hysteresis loops of a 55-nm-thick Co sample. It saturates at about 1.4 Tesla.

2.3 Making microstructures

Several types of small structures have been fabricated from these thin films using the process described in Fig. 2-2. Figure 2-9 shows a typical transport structure. The smallest wire width is 0.65 μm, while the size of contact pads is as large as 250 μm. For this structure, UV light (wavelength of 365 nm) is shone through a mask to expose reduced pattern (10:1 in a 10× stepper) in the photoresist. The resist is OCG OiR 897-71 (~1 μm). The exposed resist is developed in OCG OPD 262 (no dilution). After that, the films without resist protection are removed by Ar ion-milling. The rest of resist is removed by 1165 remover. Recipes for different resists can be seen in Appendix I.

To form good ohmic contact using wire bonding, I deposited 20 nm Ti and 200
Fig. 2 – 7 Off-axis x-ray pole figures of a hcp Co film. Diffraction peaks correspond to the in-plane (10-10) direction, and shows a 6-fold symmetry.
Fig. 2-8 Magnetization hysteresis loops measured with a superconducting quantum interference device magnetometer of a 55-nm-thick Co samples at 300 K for applied fields in plane (dashed line) and perpendicular to the film plane (solid line).
nm Au layers on top of the contact pads. It is done by lift-off process [Fig. 2-3]. For a contact aligner, Shipley 1813 resist is used. The overlay alignment accuracy is about 1 μm. After image reversal (see Appendix II), the contact layers are deposited by electron beam evaporation. Then 1165 remover is used to dissolve the resist to carry away the metals outside the contact pads.

Micron size particles have been fabricated by electron beam lithography, in which electrons are focused into small diameter spot and scanned directly onto the resist. The resist I used is SAL-603 1:1 with thinner P (3000 RPM for 280 nm), which is developed in CD-30. Figure 2-9 shows Scanning Electron Microscope image of some particles. The particle width varies from 0.25 to 2 μm. The rectangular ones have length to width ratio 3:1 or 5:1, while the other two types have additional triangular or needle-like ends. The particles are well separated by 3 times their width and four times their length.
Fig. 2 – 9  Optical micrograph of a 1 \( \mu \text{m} \) transport structure showing the Fe crystallographic orientations.

Fig. 2 – 10  SEM images of rectangular, triangular, and needle-shaped particles.
Chapter 3 Measurement Techniques

In this chapter, two of the techniques used in my research will be discussed, Magneto-optical Kerr (MOKE) hysteresis loop measurements and magnetic force microscopy (MFM). Together these techniques enabled studies of the connection between local magnetic domain configurations within elements and magnetic hysteresis of an array of such elements (such as, the coercivity, remnant magnetization and switching characteristics) – an ensemble average of the magnetic response. The physics of MOKE [3-1] will be described from a macroscopic point of view, along with the phase modulation technique used to measure MOKE signals. MFM [3-2] has been used to characterize the magnetic domain configurations and magnetization reversal process in individual magnetic elements.

3.1 Magneto-optical Kerr effect (MOKE)

3.1.1 Macroscopic description

Magneto-optical effects in ferromagnetic materials are produced by a combination of the net spin polarization that exists in the ferromagnet and spin-orbit coupling. The external magnetic field and the exchange field in ferromagnets interact with the spin magnetic moment, giving them a partial orientation. This in turn, via spin-orbit interaction, influences the orbital electron motion, with which the light electrical field couples.

MOKE is a change of the polarization state and/or the intensity of light when
it is reflected from a magnetic material. The general property that distinguishes MOKE from other magneto-optical effects in solids is that all manifestations of the Kerr effect are proportional to the Magnetization $M(T)$ and vanish at temperatures above the Curie temperature $T_C$, the temperature below which the material orders ferromagnetically.

A macroscopic description of the magneto-optic response relates measurable parameters such as reflectance, polarization changes and optical phase shifts, to general parameters that describe the media response, i.e. the conductivity tensor, dielectric tensor, or index of the refraction. Let us consider a simple case in which light is incident normal to the sample surface and parallel to the magnetization direction (Fig. 3-1). In a magnetized material, time reversal symmetry is broken. This introduces off-diagonal elements in the dielectric tensor,

$$
\begin{pmatrix}
\varepsilon_{xx} & \varepsilon_{xy} & 0 \\
-\varepsilon_{xy} & \varepsilon_{yy} & 0 \\
0 & 0 & \varepsilon_z
\end{pmatrix}
$$

where the diagonal terms are even functions of $M$ (independent of $M$ to first order) and the off-diagonal terms are odd functions of $M$ (linear dependence on $M$ to first order). All elements are complex quantities. The right circularly polarized (RCP) and left circularly polarized (LCP) light propagate differently in magnetized material, with a generally complex wave vector $k = n k_0$. The reflection coefficients, $r_+$ for RCP and $r_-$ for LCP, are

$$
r_+ = \frac{n_+ - 1}{n_+ + 1} \quad (3-1)
$$

37
\[ r = \frac{n_+ - 1}{n_+ + 1} \quad (3-2) \]

in which

\[ n_+^2 = \varepsilon_{xx} + i \varepsilon_{xy} \quad (3-3) \]

\[ n_-^2 = \varepsilon_{xx} - i \varepsilon_{xy} \quad (3-4) \]

An incident light beam linearly polarized along the x-axis (Fig. 3-1) can be written as a linear combination of a RCP and a LCP light, which can be represented by Jones vector [3-1],

\[ \left( \begin{array}{c} 1 \\ 0 \end{array} \right) = \frac{1}{2} \left[ \left( \begin{array}{c} 1 \\ i \end{array} \right) + \left( \begin{array}{c} 1 \\ -i \end{array} \right) \right] \quad (3-5) \]

where \( \left( \begin{array}{c} 1 \\ 0 \end{array} \right) \), \( \left( \begin{array}{c} 1 \\ i \end{array} \right) \), \( \left( \begin{array}{c} 1 \\ -i \end{array} \right) \) represent linear, RCP and LCP polarized light, respectively.

The Jones vector is a vector presentation of light polarization state

\[ \vec{E} = \left( \begin{array}{c} E_x \\ E_y \\ E_z \end{array} \right) \quad (3-6) \]

in which common parts are factored out to get the simplest possible expression for the vector itself. For example, the electric field, which lies in the XY plane, can be written as

\[ \left( \begin{array}{c} E_x \\ E_y \end{array} \right) = \left( \begin{array}{c} E_{0x} e^{i(kz-\omega t)} \\ E_{0y} e^{i(kz-\omega t)-\phi} \end{array} \right) = E_{0x} e^{i(kz-\omega t)} \left( \begin{array}{c} 1 \\ \frac{E_{0y}}{E_{0x}} e^{i\phi} \end{array} \right) \quad (3-7) \]
The Jones vector is
\[
\begin{pmatrix}
1 \\
\frac{E_{yy}}{E_{xx}} e^{i\phi} \\
\frac{E_{yy}}{E_{xx}}
\end{pmatrix}
\]

The reflected wave is
\[
\frac{r_r}{2} \begin{pmatrix} 1 \\ i \end{pmatrix} + \frac{r_r}{2} \begin{pmatrix} 1 \\ -i \end{pmatrix} = \frac{r_r}{2} \left\{ \begin{pmatrix} 1 \\ i \end{pmatrix} + \frac{r_r}{r_r} e^{i(\phi_+ - \phi_-)} \begin{pmatrix} 1 \\ -i \end{pmatrix} \right\}
\]  
(3-8)

This represents an elliptically polarized wave (Fig. 3-2), whose major semi-axis forms an angle \(\theta_K\) with the x-axis, denoted the Kerr angle
\[
\theta_K = \frac{\phi_+ - \phi_-}{2}
\]  
(3-9)

while the ratio of semiaxis is
\[
\varepsilon_K = \frac{1 - \frac{r_r}{r_+}}{1 + \frac{r_r}{r_+}}
\]  
(3-10)

and is called the Kerr ellipticity.

Defining \(R\) such that
\[
\left| \frac{r_r}{r_+} \right| = e^R
\]  
(3-11)

We have
\[
\varepsilon_K = \frac{1 - e^R}{1 + e^R} = -\tanh \left( \frac{R}{2} \right)
\]  
(3-12)

From equation (3-1) and (3-2), one gets:
Fig. 3 - 1 The polar Kerr effect geometry. Incident light is normal to the sample surface and parallel to the magnetization direction. The incident light is linearly polarized with the polarization along the X direction.

Fig. 3 - 2 Elliptically polarized state of reflected light. Dashed circles show the LCP and RCP components. When they are parallel, their moduli add up to give the major semi-axis and form with the X-axis an angle $(\phi^- - \phi^+)/2$. When they are antiparallel, their moduli subtract to give the minor semi-axis.
\[
\frac{n_- - n_+}{n.n_+ - 1} = \frac{r_- - r_+}{r_+ + r_-} = \frac{1 - \frac{R}{r_+} e^{i(\phi_+ - \phi_-)}}{1 + \frac{R}{r_-} e^{i(\phi_- - \phi_+)}} = \frac{1 - e^{R_+ - i(\phi_+ - \phi_-)}}{1 + e^{R_+ - i(\phi_- - \phi_+)}} = -\tanh\left(\frac{R}{2} + i\frac{\phi_- - \phi_+}{2}\right) = -\frac{R}{2} - i\frac{\phi_- - \phi_+}{2} = -\varepsilon_k - i\theta_k
\]

(3-13)

In general both \( R \) and \((\phi_- - \phi_+)\) are much smaller than 1. To a good approximation

\[ n_- + n_+ = 2\tilde{n} \quad \quad n.n_+ \approx \tilde{n}^2 = \varepsilon_{\alpha\alpha} \quad \quad (3-14) \]

We then obtain

\[
\theta_k = -\text{Im}\left[\frac{n_- - n_+}{n.n_+ - 1}\right] = -\text{Im}\left[\frac{n_-^2 - n_+^2}{(n.n_- - 1)(n_- + n_+)}\right] = -\text{Im}\left[\frac{\varepsilon_{\alpha\alpha} + i\varepsilon_{\alpha\gamma} - (\varepsilon_{\alpha\alpha} - i\varepsilon_{\gamma\gamma})}{2\tilde{n}(\tilde{n}^2 - 1)}\right]
\]

\( = -\text{Im}\left[\frac{i\varepsilon_{\gamma\gamma}}{n(\varepsilon_{\alpha\alpha} - 1)}\right] = -\text{Im}\left[\frac{i\varepsilon_{\gamma\gamma}}{n(\varepsilon_{\alpha\alpha} - 1)}\right] = -\frac{\varepsilon_{\gamma\gamma}}{n(\varepsilon_{\alpha\alpha} - 1)} \right] = -\frac{\alpha}{n(\varepsilon_{\alpha\alpha} - 1)} M
\]

(3-15)

Similarly,

\[
\varepsilon_k = \frac{\varepsilon_{\gamma\gamma}}{(\varepsilon_{\alpha\alpha} - 1)n} = \frac{\alpha}{(\varepsilon_{\alpha\alpha} - 1)n} M
\]

(3-16)

The above formulae show that the Kerr rotation and ellipticity are proportional to the magnetization \( M \).

There are two other Kerr configurations shown in Fig. 3-3. In longitudinal case, the magnetization lies in the plane of incidence and parallel to the film surface.

The magneto-optic effect consists of an \( M \)-dependent change in elliptical polarization of the reflected beam. In the transverse case, the magnetization is perpendicular to the plane of incidence and the component confined to the plane of
Fig. 3-3 a) longitudinal magneto-optic Kerr effect geometry, b) transverse magneto-optic Kerr effect geometry.
incidence of the reflected beam exhibits an $M$-dependent change in intensity.

3.1.2 Phase modulated MOKE measurement

3.1.2.1 The phase modulator

The magneto-optic Kerr effect has provided an important means of probing a broad range of thin film magnetic properties. As such, various techniques have been developed to make high sensitivity measurements of the effect. One advanced technique is to use a photoelastic modulator [3-3]. The photoelastic modulator is an instrument used to modulate (at fixed frequency) the polarization of a beam of light. It is based on the photoelastic effect, in which a mechanically stressed sample exhibits birefringence proportional to the resulting strain. It consists of a rectangular bar of a suitable transparent material (fused silica for example) attached to a piezoelectric transducer. By applying an ac voltage, the transducer is periodically strained by piezo-electric effect. This strain is elastically transmitted to the bar of fused silica, that assumes a strain-induced periodic birefringence.

3.1.2.2 Principles of the technique

A schematic diagram showing the principle of the technique is given in Fig. 3-4. $P$ denotes the polarizer, the polarizing axis of which makes angle of 45° with the $x$-axis. $Mo$ denotes the photoelastic modulator, the optical axis of which is in the $x$-direction. The light passing through $Mo$ receives a periodically varying retardation $\delta$, which is described by

$$\delta = \delta_0 \sin 2\pi ft$$  \hspace{1cm} (3-17)
Fig. 3 - 4  Schematic diagram showing essentials of the measuring technique.

P: polarizer, Mo: modulator, A: analyzer, D: detector. The polarizing angle of $p$ makes $45^\circ$ with the $x$-axis. The polarizing angle $A$ is assumed to be $\theta$. 
Modulated light is reflected by a sample S, then passes through an analyzer A, the polarizing axis of which makes an angle $\theta$ with x-axis. Finally, the light is detected by a photodetector D.

It has been shown [3-3] that the intensity $I$ of the light detected by D is given by

$$I = \frac{eE^2}{16\pi} \left[ R + \frac{\Delta R}{2} \sin\delta + R\sin(\Delta\phi + 2\theta)\cos\delta \right] \tag{3-18}$$

where $R$, $\Delta R$ and $\Delta\phi$ are defined as

$$R = \frac{1}{2} \left( |r_+|^2 + |r_-|^2 \right), \quad \Delta R = |r_+|^2 - |r_-|^2, \quad \Delta\phi = \phi_+ - \phi_- \tag{3-19}$$

The magneto-optical parameters-Kerr rotation $\theta_K$ and Kerr ellipticity $\epsilon_K$ can be expressed by the following equations:

$$\theta_K = -\frac{1}{2} \Delta\phi \quad \quad \epsilon_K = \frac{1}{4} \left( \frac{\Delta R}{R} \right) \tag{3-20}$$

By using the expansion formulae

$$\sin(\delta_0 \sin 2\pi ft) = 2J_1(\delta_0) \sin 2\pi ft + ... \tag{3-21}$$

$$\cos\delta = \cos(\delta_0 \sin 2\pi ft) = J_0(\delta_0) + 2J_2(\delta_0) \sin 4\pi ft + ... \tag{3-22}$$

where $J_n(x)$ is an n-th order Bessel function, we obtain

$$I = I(0) + I(f) \sin 2\pi ft + I(2f) \sin 4\pi ft + ... \tag{3-23}$$

in which

$$I(0) = I_0 R \tag{3-24}$$

$$I(f) = I_0 \Delta R J_1(\delta_0) \tag{3-25}$$
\[ I(2f) = 2I_o R J_2(\delta_o) \sin(\Delta\phi + 2\theta) \] (3-26)

The output voltage of the detected D therefore consists of three components: a dc component, an f-component and a 2f component. If \( \theta = 0 \), then

\[ \frac{I(2f)}{I(0)} \approx 2J_2(\delta_o)\Delta\phi = -4J_2(\delta_o)\theta \kappa \] (3-27)

\[ \frac{I(f)}{I(0)} \approx J_1(\delta_o)\frac{\Delta R}{R} = 4J_2(\delta_o)\varepsilon \kappa \] (3-28)

Therefore we are able to determine the values of the Kerr rotation and Kerr ellipticity by detecting the f and 2f components, respectively.

3.1.2.3 Experimental set-up

In our experiment (Fig. 3-5), we use a diode laser as light source. The wavelength is 635 nm at 25 °C. The photoelastic modulator (Hinds Instruments, PEM-90) operates at 50 kHz. The reflected light is detected by a photodiode detector/preamplifier, then the output signal is fed into a lock-in amplifier. The frequency component at 100 kHz is detected, which is a measure of the Kerr rotation. A magnetic field is applied parallel to the sample surface and in the plane of incident light, which provides a longitudinal Kerr-effect condition.

3.2 AFM/MFM

MOKE can be used to determine the ensemble average magnetic properties of an array of small magnetic elements. While MFM has the capability to resolve the
Fig. 3 – 5 MOKE measurement. 1. Diode laser; 2. Polarizer, the polarizing angle is $45^\circ$ to the horizontal direction; 3. Photoelastic modulator, operating at 50 kHz; 4. Sample; 5. Analyzer, whose polarizing angle is parallel to the horizontal; 6. Photodioade detecot/preamplier; 7. Lock-in amplifier; 8. Computer; 9. Electromagnet.
magnetic state of individual particles. The basic concept of MFM is illustrated in Fig. 3-6. Stray magnetic field emanating from the surface of a sample generates a force on a magnetic tip, which is attached to a flexible cantilever. A sensitive deflection sensor is used to detect cantilever motion and hence the force (or force gradient). To form an image, the strength of the sample-tip interaction is mapped as a function of position on the sample.

The sensing element of the commercial MFM (DI 3000) is a single crystal silicon cantilever whose tip is coated with a magnetic film. One type of tip is coated with a CoCr film of 50 nm thickness that has an effective moment of $\sim 10^{12}$ emu and coercivity of $\sim 500$ Oe. To obtain a magnetic force image, first the topography of the sample is scanned in tapping mode. In this mode, the cantilever is oscillated at its resonance frequency (60-100 kHz) with a bimoph piezo and makes intermittent contact with the sample. At contact, the van der Waals interaction between sample and tip is dominant, making the measured signal very sensitive to topographic effects. The amplitude of the oscillation is proportional to the height of the tip from sample. By adjusting the tip height to maintain constant amplitude and force on the sample, the topography of the samples is imaged. The same line of the sample is then scanned again, but at a constant separation between the tip and sample, using the topographic information from the first scan as a reference. Thus, the tip is now only sensitive to long range force such as a magnetic interaction. A spatial gradient in the magnetic force between the tip and the samples produces a shift of a few Hz in the resonance frequency of the cantilever. For a small force gradient the associated phase
Fig. 3 – 6  Basic concept of magnetic force microscopy. A magnetic tip attached to a flexible cantilever is used to detect the magnetic field emanating from a sample.
shift or amplitude change is proportional to the magnitude of the force gradient and thus, the magnetic force image is obtained.

The influence of a magnetic interaction is measured using the principle of force gradient detection. Consider the magnetic tip as a point dipole \( \vec{m} \), and the field from sample is \( \vec{H} \), the total potential energy is

\[
U = \frac{1}{2} k z^2 - \vec{m} \cdot \vec{H}
\]  

(3-29)

The cantilever is usually parallel to the sample (Fig. 3-6), so it is mostly sensitive to the vertical force

\[
F_z = -\frac{\partial U}{\partial z} = -kz + \vec{m} \cdot \frac{\partial \vec{H}}{\partial z}
\]  

(3-30)

The effective spring constant is proportional to the force gradient

\[
k_{\text{eff}} = -\frac{\partial F_z}{\partial z} = k - \vec{m} \cdot \frac{\partial^2 \vec{H}}{\partial z^2}
\]  

(3-31)

The magnetic force gradient is generally much smaller than the spring constant of the cantilever, therefore, the change of resonance frequency is proportional to the magnetic force gradient.

\[
\omega_0 \propto \sqrt{k_{\text{eff}}} = \sqrt{k \left(1 - \frac{1}{2k} \vec{m} \cdot \frac{\partial^2 \vec{H}}{\partial z^2}\right)}
\]  

(3-32)

This also shows that the component of stray field, which is sensed, depends on the orientation of the tip moment. In my experiments the tip is always magnetized vertically.

Since MFM is essentially measuring the derivative of the magnetic field, the
interpretation of the MFM images is not trivial. Nevertheless, MFM will highlight magnetic poles at the boundaries (-n\*M) of the samples and magnetic charges associated with domain walls (\(\nabla \bullet M\)), from which the stray field is emanating (Fig. 3-7).
Fig. 3 – 7  A MFM image observed in a Fe wire. Domain walls show strong bright
and dark contrast, while inside each domain the contrast is weaker and uniform.
Chapter 4 Domain Wall Scattering

The low field magnetoresistance (MR) of a ferromagnetic metal depends in a detailed and usually complex manner on its magnetic domain structure. An understanding of the interplay between transport and magnetic properties is important to the miniaturization of magnetic devices to nanoscale dimensions, such as those based on giant magnetoresistance (GMR), and to the interpretation of transport results on magnetic nanostructures. A recent focus has been to develop model thin film structures to study domain wall (DW) scattering on conduction electrons. A magnetic configuration, which consists of oppositely magnetized ferromagnetic domains separated by domain walls, closely resembles that of a magnetic trilayer, i.e. two regions of oppositely pointing magnetization separated by a thin non-magnetic interlayer (Fig. 4-1). So why does not a chemically homogeneous thin ferromagnetic film exhibit GMR? To answer this question, we have studied DW scattering effects on three different materials: epitaxial bcc Fe, hcp Co, and L1₀ FePt, with progressively high magnetic anisotropy, and thus thinner DWs. Several possible mechanisms of MR due to DWs will also be discussed.

4.1 Intrinsic domain wall effects

In the 1970s Cabrera and Falicov considered two mechanisms by which the domain walls produce additional scattering [4-1]. The first mechanism comes from
Fig. 4-1 Schematic illustration of the similarity between the magnetic geometry of a GMR trilayer and a ferromagnetic domain wall, a) an antiferromagnetic aligned trilayer and the presence of a domain wall, b) the trilayer in its ferromagnetically aligned state and the absence of a domain wall in the ferromagnet.
the reflection of incoming electron waves from the ferromagnetically ordered domains as they entered the twisted spin structure of a wall. This effect is only sizeable when the spin-up and down density of states are considerably different and the DW is very narrow. Berger [4-2] showed that, since the conduction electron wavelength (0.15 nm) is much shorter than the domain wall length (10 nm), the electron spin follows the local magnetization adiabatically, and the reflection is negligible, unless the walls are so abrupt to be of monolayer thickness. The second mechanism is due the zig-zagging character of the electron orbits when going between up- and down-regions of the magnetization.

While more recently, Gregg et al [4-3] reported a large MR effects due to DWs even at room temperature in simple ferromagnetic films. They proposed that “the key to electrical scattering by domain walls is how well the precessional behavior of the carrier spin allows it to track the changing local exchange field direction as it traverses the wall”. The momentum scattering rate of polarized electrons is a linear function of the cosine of the angular deviation between the electron spin and the scattering center local magnetic moment. They found the magnetoresistance effect due to a DW is

\[
\frac{\Delta R}{R} = \frac{2p}{(1-p)^2} \left( \frac{\hbar v_F E_{ex}}{E_{ss}} \right)^2 \frac{1}{\delta_w d_s}
\]

(4-1)

where \( p \) is the spin-dependent scattering ratio, \( v_F \) is the Fermi velocity, \( E_{ex} \) is the exchange energy, and \( \delta_w, d_s \) are magnetic domain wall width and domain size [4-4].
Independently, a new physical mechanism has been proposed to explain these observations, which invokes the two channel model of conduction in ferromagnets and spin dependent electron scattering [4-5]. Within this model, DWs increase resistivity because they mix the minority and majority spin channels and thus partially eliminate the short circuit provided by the lower resistivity spin channel in the magnetically homogeneous ferromagnet. It predicts that when current is parallel to the domain walls (CW), the magnetoresistance ratio \( R_{cw} \) is

\[
R_{cw} = \frac{\rho_{cw} - \rho_0}{\rho_0} = \frac{\xi^2}{5} \left( \frac{\rho_{c+} - \rho_{c0}}{\rho_0 \rho_{c0}^*} \right)^2
\]

(4-2)

And the magnetoresistance ratio \( R_{cpw} \) for current perpendicular to domain walls (CPW) is

\[
R_{cpw} = \frac{\rho_{cpw} - \rho_0}{\rho_0} = \frac{\xi^2}{5} \left( \frac{\rho_{cp0} - \rho_{c0}^*}{\rho_0 \rho_{c0}^*} \right)^2 \left( 3 + \frac{10\sqrt{\rho_{cp0}^*}}{\rho_0 + \rho_{c0}^*} \right)
\]

(4-3)

here \( \xi = \hbar v_f/(J\delta) \) is a measure of the non-adiabaticity (\( \xi = 0 \) is an adiabatic crossing). \( \rho_{c0}^* \) is the resistivity for spin s of the ferromagnet.

The above experiments and theories show that domain walls increase resistance. Whereas in other experiments [4-6], which use MR to study domain wall dynamics, it is found that nucleation of a domain wall appears to decrease resistance. A novel theoretical explanation [4-7] has been proposed in which DWs destroyed the electron coherence for weak localization at low temperature, leading to such a negative DW contribution to the resistivity. Bratass and Bauer [4-8] also showed that DW resistance of diffusive ferromagnets could be negative as well as positive when...
the electronic structure of the domain wall is taken into account semiclassically. A modified magnetization causes a change of resistivity

$$\delta \rho = -\rho_0^2 \frac{g^2}{m} \delta n \left( \tau_+ - \tau_- \right)$$  \hspace{1cm} (4-4)

Where $\rho_0$ is the resistivity of a single-domain ferromagnet, $\delta n$ is the change of the density of spin-up or spin-down electrons, and $\tau_+$ is the scattering relaxation time for the spin-up or spin-down electrons. So it depends on the spin-dependent relaxation times, thus the types of impurities that are present in the material.

4.2 Extrinsic domain wall effects

The problem to find the intrinsic domain wall effect is complicated by the extrinsic domain wall contributions to the resistance, such as anisotropic magnetoresistance (AMR). The resistance of ferromagnet depends on the details of domain configuration and may be larger in a multi-domain state than that of a single domain state. Berger [4-2] has also discussed a domain mechanism, which can enhance resistivity based on the Hall effect. The Hall effect leads to current deflection near a domain wall (Fig. 4-2), due to the angle between the electric field and the current in domains. Since the Hall angle changes sign in alternating magnetization domains, current will zigzag through the sample and the resistivity is predicted to be increased of order $(\rho_\alpha/\rho_\omega)^2$, the Hall angle squared.

57
Fig. 4-2  Current distribution in the case of stripe domains normal to the average current.
4.3 Domain wall scattering in Fe films

4.3.1 Fabrication and experiments

Bcc (110) Fe films with thickness 25, 50, 100, and 200 nm have been grown using the method described in Chapter 2. These films show a strain-induced in-plane uniaxial anisotropy with the easy axis along the [001] direction, in addition to the cubic anisotropy which is characteristic for bulk bcc Fe. A typical 100-nm-thick Fe film had a residual resistivity ration of 30, and the residual resistivity is 0.2 $\mu\Omega$cm. These indicate the high crystalline quality of these films.

Using optical projection lithography techniques and ion milling Fe wires of 0.65 $\mu$m to 20 $\mu$m linewidth were fabricated with the magnetic easy axis perpendicular to the wire axis [Fig. 2-9]. The wire length is about 200 $\mu$m. Fe wire arrays with same widths were also patterned to investigate the magnetic properties and the magnetization reversal process via SQUID and longitudinal Kerr measurements. Magnetic force microscope has been used to image remnant domain structures and magnetization reversal process under external field. The MR measurements were performed in a variable temperature high field cryostat with in situ (low temperature) sample rotation capabilities. The applied field was in the plane of the film and oriented either longitudinal (||) or transverse (⊥) to the wire axis. The resistivities have been measured using a 4 probe ac (−10 Hz) bridge technique with low currents (10 $\mu$A).
4.3.2 Magnetic characterization

The competing magnetostatic, magnetocrystalline, and exchange interactions result in a regularly spaced stripe domain configuration with flux closure domains. Varying the wire width changes the ratio of these energies and hence the domain size. Figure 4-3 shows MFM images of the domain configurations of 2.0 μm, 5.0 μm, and 20 μm wires. Before imaging the wires were magnetized to saturation with a magnetic field transverse (Fig. 4-3 a, c, and e) or longitudinal (Fig. 4-3 b, d, and f) to the wire axis. With the increasing wire width, domain sizes become larger. After longitudinal saturation, the average domain size in 20 μm wire is about 6 μm, while the average domain size is 0.45 μm in 2.0 μm wire. The flux closure domain configurations at the boundaries of the wires also change with the wire width. For broad Fe wires (≥ 10 μm) so-called Landau-Lifshitz pattern with spike domains is present (Fig. 4-3 e, f, and a magnification showed in Fig. 4-4), compared to the simple triangular closure domain in narrower wires (Fig. 4-3 a, b, c, and d). The magnetic domain configurations are strongly affected by the magnetic history of the samples. Usually the domain width is larger after transverse saturation than that after longitudinal saturation. For example, in the transverse case 2 μm wide wires show a mean stripe domain length of 1.6 μm, which is much larger than 0.4 μm in the longitudinal case. Obviously, the domain width does not obey the simple square root law, shown in Equation (1-5). What we have observed are metastable states of the system. There are energy barriers to overcome to reach the equilibrium state. These

60
Fig. 4-3  MFM images in zero applied field of (a, b) 2 μm, (c, d) 5 μm, and (e, f) 20 μm linewidth Fe wires. Images in the left column were taken after magnetic saturation transverse to the wire’s long axis, while those in the right column were taken after longitudinal saturation. The dashed lines in (a) illustrates the flux closure domain configurations observed.
Fig. 4-4  A magnified MFM image of Landau-Lifshitz pattern observed at wire edge.
observed domain structures at $H=0$ are stable over observation times of at least several hours showing that the DWs are strongly pinned at room temperature.

Since current is directed along the wire, there are domains with magnetization oriented both parallel and perpendicular to the current density. In order to estimate the MR contribution due to resistivity anisotropy the volume fraction contributions (with $M//J$) has been determined from MFM images. Fig. 4-5 shows this fraction (labeled $\gamma$), together with domain size, after magnetic saturation in either the transverse or longitudinal direction. In narrow wires, $\gamma$ can be as high as 40%.

Figure 4-6 shows that we can continuously vary DW density in a single wire by applying a saturation field in different directions. The images were taken in zero applied field after the wire was magnetized to saturation a) longitudinal ($\Theta=0^\circ$), b) $\Theta=60^\circ$, c) $\Theta=72^\circ$, and d) transverse ($\Theta=90^\circ$) to the wire axis. Figure 4-7 shows the continuous increase of the stripe domain width from 0.45 $\mu$m to 1.8 $\mu$m as function of the angle $\Theta$ between the applied field and the wire axis.

In order to understand the large difference in stripe domain width between the longitudinal and transverse geometry MFM measurements were performed just below the saturation fields for these configurations. In the longitudinal case shown in Fig. 4-8 a), besides the triangle shaped flux closure domains with the magnetization parallel to the applied field ($H=0.013$ T) stripe domains of alternating magnetization direction perpendicular to the field are nucleated. A further reduction of the applied
Fig. 4-5 The right-hand axis displays the domain width versus Fe linewidth in zero applied field after transverse (open squares) and longitudinal (open circles) magnetic saturation. The left-hand axis shows the volume fraction of closure domains $\gamma$ as function of the linewidth, again, after transverse (solid squares) and longitudinal (solid circles) magnetization.
Fig. 4-6  The MFM images of the stripe domain pattern of a 2 µm Fe wire in zero applied field show a strong dependence of the previous saturation direction. Before performing the MFM imaging the wire was magnetized to saturation (a) longitudinal (θ=0°), (b) θ=60°, (c) θ=72°, and (d) transverse (θ=90°), with respect to the wire axis.
Fig. 4-7  The domain width in zero-applied field determined from MFM images increases continuously as function of the previous saturation direction labeled by the angle $\Theta$ between the wire axis and the applied field.
field does not lead to a change of the stripe domain, only a contraction of the volume of the flux closure domains occurs via DW movement. In contrast, the transverse case just below the saturation field (H=0.03 T) shown in Fig. 4-8 b) is dominated by a stripe domain configuration of alternating magnetization direction, where the anti-parallel oriented domains are smaller in volume. As in the longitudinal case, a further reduction of the field leads to a DW movement but not further nucleation of DWs. In both instances the nucleations of DWs is associated with the formation of domains oriented perpendicular to the applied magnetic field. In the longitudinal case, these domains are along the magnetocrystalline easy axis. However, in the transverse they are perpendicular to the magnetocrystalline easy axis. Thus, qualitatively, the lower DW density in the transverse case is due to the larger magnetocrystalline energy associated with the magnetic state after DW nucleation.

Figure 4-9 shows that there is a transition of the DW type as a function of film thickness. In 25 nm thick wire, there is zigzag Neel-like DWs (Fig. 4-9 a), in which the magnetization rotates in the plane. In contrast, 50 nm and 100 nm thick wires show Bloch walls, in which the magnetization rotates out of plane. The transition from a Neel to a Bloch wall occurs when the thickness of the film is in the range of the DW width and is driven by the magnetostatic energy of the DW. If the wire is thinner than the DW width, the shape anisotropy of the DW favors a Neel-like DW. Following the same arguments, if the film thickness is greater than the DW width shape anisotropy favors Bloch-like DWs. Additionally, the subdivision of the
Fig. 4-8 MFM images of the domain configurations of 2 μm Fe wire in the presence of (a) a longitudinal field of $H=0.013$ T, and (b) a transverse field of $H=0.03$ T, which are both just below the saturation field of the given orientation.
Fig. 4-9 MFM images in zero applied field of (a) 25 nm, (b) 50 nm, (c) 100 nm, and (c) 200 nm thick Fe wire of 2 μm width after longitudinal saturation.
DWs into sections of opposite chirality lowers the magnetostatic energy, as
alternating black or white contrast seen in Fig. 4-9 (a) and partly in (b). In even
thicker films (e.g. 200 nm), DWs become canted to bring opposite magnetic closer,
thus further reducing the magnetostatic energy.

4.3.3 Transport properties

Figure 4-10 shows a representative MR result on a 2 μm wide wire at both (a)
high (270 K) and (b) low temperature (1.5 k). First consider the MR characteristics at
270 K. For field transverse to the wire, and thus parallel to the magnetic easy axis,
the MR is negative (i.e. the MR decreases with increasing field). While in the
longitudinal geometry, the low field MR is positive, and hysteresis is observed. At
fields greater than the magnetic saturation fields (H_{ys} = 0.035 T and H_{sx} = 0.085 T)
the resistivity is essentially independent of magnetic field and \( \rho_L(H) > \rho_T(H) \). At 1.5
K these characteristics change significantly. The transverse MR is now positive at
low field, while the longitudinal MR is more complex. Above the saturation field the
MR is large and positive, and the resistivity anisotropy is reversed, \( \rho_L(H) < \rho_T(H) \).

There are two important sources of resistivity, which must be considered to
interpret this transport data. The first has its origins in the spin-orbit coupling and is
know as AMR – the resistivity extrapolated back to zero internal field (B = 0)
depends on the relative angle between M and J. The second effect is due to the
ordinary (Lorentz) MR, and is also in general anisotropic (i.e. dependent on the
relative orientation of J and B). As Fe has a large magnetization and hence a large

70
Fig. 4-10 (a) MR data at 270 K of a 2 μm wire in the transverse and longitudinal field geometries [$\rho_z(H = 0, 270 \text{ K}) = 14.7 \mu\Omega\text{cm}$]. (b) MR at 1.5 K [$\rho_z(H = 0, 1.5 \text{ K}) = 0.74 \mu\Omega\text{cm}$].
internal magnetic field \((4\pi M = 2.2 \, \text{T})\) both factors are of importance. The resistivity of domains parallel and perpendicular to the current direction can be written as

\[
\rho_\parallel (B, T) = \rho_\parallel (0, T) \left[ 1 + F_\parallel (B/\rho_\parallel (0, T)) \right] \tag{4-5}
\]

\[
\rho_\perp (B, T) = \rho_\perp (0, T) \left[ 1 + F_\perp (B/\rho_\perp (0, T)) \right] \tag{4-6}
\]

here \(B\) is the internal field in the ferromagnet, \(B = 4\pi M + H - H_d\), with \(H\) the applied field and \(H_d\) the demagnetization field. The AMR is equal to \([\rho_\parallel (0, T) - \rho_\perp (0, T)]/\rho(0, T)\), where \(\rho(0, T)\) is the average resistivity. The function \(F\) is known as Kohler function and parametrizes the ordinary magnetoresistance for longitudinal and transverse field geometries in terms of \(B/\rho \sim \omega_c \tau\), the cyclotron frequency times the relaxation time. These scaling functions have been determined experimentally by performing MR measurements to large fields (6 T) as a function of temperature, as described in [4-9]. Figure 4-11 shows an example of high-field MR data at different temperatures from 1.5 K to 150 K. At lower temperature, the Lorentz MR effect is stronger, due to the increased electron scattering time. The following form of Kohler's rule is used to analyzing the data

\[
\kappa(T) \rho(B, T) = \rho^* \left[ 1 + F_\kappa (\kappa(T) B/\rho^*) \right] = G(\kappa(T) B) \tag{4-7}
\]

where the parameter \(\kappa(T) = \rho^*/\rho(0, T)\). \(\rho^*\) is arbitrary constant for the temperature series. In our analysis, \(\rho^*\) is usually first extrapolated from MR curve at \(T = 40 \, \text{K}\). In this form, Kohler's rule implies \(\kappa \rho\) is a function only of \(\kappa B\). In practice, \(\kappa\) is chosen to be one for \(T = 40 \, \text{K}\) and values of \(\kappa\) for other temperatures are found by normalizing the data to fall on one smooth curve. Note that unique values of \(\kappa\) are
found. The form of the function, $G(\kappa B)$ can then be mapped over large ranges of the argument $\kappa B$, facilitating an extrapolation of $\kappa\rho$ to the value $\rho^*$ at zero field. From this value and the values of $\kappa(T)$ the resistivities $\rho(0,T)$ can be deduced. Figure 4-12 shows such a scaled curves for 2 $\mu$m Fe wire. The inset displays both $\rho_\perp(0,T)$ and $\rho_\parallel(0,T)$ which result from this scaling analysis. We found $\rho(0,T) \sim aT^2$ with $a = 3 \times 10^{-4} \mu\Omega$cm/K$^2$, as typically observed in 3d elemental ferromagnets [4-10]. The AMR is $\sim 4 \times 10^{-3}$ above 80 K and decreases below this temperature. The reversal of the resistivity anisotropy at low temperatures [$\rho_\perp(H) > \rho_\parallel(H)$, Fig. 4-10] is thus mainly a consequence of the increasing importance of the Lorentz MR which is opposite in sign to the AMR.

As in all ferromagnetic materials the resistivity anisotropy is a source of low field MR. An applied field changes the domain configurations, and domains with magnetization parallel and perpendicular to the current direction have different resistivity. Hence, this low field MR simply reflects the domain geometries during magnetization.

There are thus two ways to estimate the DW contribution to the resistivity. The first is to perform MR measurements at the temperature at which this resistivity anisotropy at $H=0$ vanishes. Since the AMR and Lorentz MR contributions to the resistivity anisotropy are of opposite sign, there will be a temperature at which $\rho_\parallel(H = 0, T_{\text{comp}}) = \rho_\perp(H = 0, T_{\text{comp}})$, which we denote the compensation temperature, $T_{\text{comp}}$. This occurs at 65.5 K and MR results are shown in Fig. 4-13 for a 2 $\mu$m wire. The
Fig. 4-11 Temperature dependence of MR above the saturation field in 2 μm Fe wire.
Fig. 4-12 Scaling plot of the transverse and longitudinal MR above magnetic saturation for a 2 μm Fe wire in the form $\rho(B)/\rho(B = 0)$ versus $B/\rho(B = 0)$ at temperatures of (open squares) 1.5 K, (solid triangles down) 40 K, (open circles) 60 K, (solid circles) 80 K, (solid triangles up) 100 K, (solid diamonds) 125 K, and (open diamonds) 150 K. The inset shows the scaling parameters $\rho_{\perp}(B = 0)$ and $\rho_{\parallel}(B = 0)$ as a function of temperature on a log-log plot.
Fig. 4-13 MR of 2 μm Fe wire at 65.5 K. The extrapolation of the high field MR data in transverse (dotted line) and longitudinal (solid line) geometry shows that \( \rho_{||}(H=0) = \rho_{\perp}(H=0) \). The resistivity with walls present, \( \rho(H=0) \), is smaller than this extrapolation and indicates that DWs lower the wire resistivity. The inset shows this negative DW contribution as a function of linewidth at this compensation temperature in the longitudinal geometry.
absence of in-plane resistivity anisotropy at $H=0$ at this temperature is suggested by extrapolation from the high-field resistivity above magnetic saturation for transverse (solid line) and longitudinal (dotted line) applied fields. The effect of domain configurations on the MR should vanish. However, the measured resistivities in both field geometries at $H = 0$ are lower than the extrapolated value and are correlated with the DW densities after longitudinal (Fig. 4-3 a) and transverse (Fig. 4-3 b) saturation. We attribute this difference to DW-MR. Here, the higher DW density state gives rise to a lower resistivity. The magnitude of the effect also decreases systematically with increasing wire width (Fig. 4-13 inset), and hence decreasing DW density (Fig. 4-3).

A more definitive correlation between domain configurations measured at room temperature using an MFM and low temperature MR measurements has been established. To do this, we warm the sample to room temperature, cycle the magnetic field to establish a known $H = 0$ magnetic state, and cool. The resistivity at $H = 0$ and the MR at low temperatures are unchanged for these samples in both longitudinal and transverse measurement geometries. Figure 4-14 shows an example. This is a strong evidence that domain structure is not affected by temperature in this range.

The DW MR dependence on DW density can be studied more elegantly by varying the DW density in a single 2 μm wire (see Fig. 4-13). In this manner effects associated with changing the wire cross section or impurity concentration can definitely be isolated from those associated with domain configurations and DWs. To achieve that, different DW densities were established by saturating the wires in
Fig. 4-14 Transverse MR at \( T = 65.5 \) K for a 1.5 \( \mu \)m Fe wire after the wire warmed up to room temperature and establish a known transverse magnetic state. The thicker curve shows the initial scan after annealing.
an in-plane direction tilted by $\Theta=10^\circ$ and $\Theta=30^\circ$ in respect to the wire axis. As discussed above for $\Theta=10^\circ$ the DW density will be clearly higher than for $\Theta=30^\circ$. After saturating and before performing the MR measurement the wire has been oriented in the transverse geometry at $H = 0$. The low field MR data (Fig. 4-15) for $\Theta=10^\circ$ and $\Theta=30^\circ$ show a continuous decrease of the resistivity with respect to the transverse case. This kind of measurements has been done for a large set of different angles as plotted in the inset of Fig. 4-15. Here, a continuous decrease of $-\rho_{\perp}/\rho(H=0)$ between the longitudinal ($\Theta=0^\circ$) and the transverse ($\Theta=90^\circ$) orientation is observed.

The temperature dependence of the DW contribution to the resistivity is estimated as follows. As resistivity anisotropy is small, the current density in each domain is to a good approximation independent of the precise domain configurations. The domain size is also greater than the mean free path ($\sim 200$ nm at 1.5 K in a 100 nm thick film). Thus we can write the effective resistivity in the $H=0$ magnetic state as:

$$\rho_{\text{eff}}(H=0,T) = \gamma \rho_{\parallel}(B_{T}, T) + (1 - \gamma) \rho_{\perp}(B_{T}, T) \quad (4-8)$$

where $\gamma$ is the volume fraction of domains oriented longitudinal and $B_{T}$ is the field internal to these domains. We determine $\rho_{\perp}(B, T)$ and $\rho_{\parallel}(B, T)$ by extrapolation of the MR data above saturation to $H = 0$ (again, as indicated by the dashed and solid lines in Fig. 4-13). The effective resistivity at $H = 0$ is estimated with the MFM measurements of $\gamma$. Deviations from this $\rho_{d} = \rho(H=0) - \rho_{\text{eff}}(H=0)$, i.e. the measured $H = 0$ resistivity minus this effective resistivity, are negative and depend systematically
Fig. 4-15 MR of a 2 μm Fe wire [$\rho(H = 0, 65.5 \text{ K}) = 1.6 \mu\Omega\text{cm}]$ measured at the compensation temperature of 65.5 K. Also included are the MR data measured in transverse geometry from $H = 0$ to saturation after magnetizing the wire transverse ($\Theta = 90^\circ$), $\Theta = 30^\circ$, $\Theta = 10^\circ$, and longitudinal ($\Theta = 0^\circ$), with respect to the wire axis. The inset shows the reduction of resistivity $-\rho_d/\rho(H=0)$ due to the presence of DWs as a function of $\Theta$. 
Fig. 4-16 The DW contribution to MR as a function of temperature of a 2 μm Fe wire.
on domain wall density. They approach 1.3% of the wire resistivity at 1.5 K for a 2 
μm wire. We also find that the magnitude of $\rho_d$ decreases with temperature 
approaching zero at ~ 80 K (Fig. 4-16). This enhancement of the conductivity 
vanishes at ~ 80 K for all the wire widths investigated.

The DW MR is further studied in wires of varied thickness and, hence, DW 
spin structures. Fig. 4-17 b) shows the MR of a 200 nm thick wire. The longitudinal 
MR is now negative at low fields and positive at higher fields. This low field 
negative MR is associated with a change in the magnetization reversal mode in 
thicker wires. For 200 nm thick film stripe domain configurations are observed (inset 
of Fig. 4-17 b). This image suggests that the magnetization has rotated out of the 
plane along \{100\} easy direction 45 degrees to the surface normal. The reduction in 
internal field (due to the strong demagnetization fields perpendicular to the film 
plane) is responsible for the low field drop in resistivity. At its compensation 
temperature 64.5 K, the measured $H = 0$ resistivity is observed to equal to the $H = 0$ 
resistivity extrapolated from the high field MR, thus $\rho_d$ vanishes at this temperature 
for 200 nm thick wire.

Varying the film thickness changes both the DW spin structure and the 
relative importance of bulk and thin film surface scattering on the resistivity. Table 
4-1 summarizes the results of a systematic study of the effect of film thickness on $\rho_d$ 
and the film resistivity. It is seen that even in thin layers (25 nm films), in which the 
DWs are considerably broader and Neel-like; there is a large anomalous negative 
DW contribution. The magnitude of DW interface resistance (which is negative) and
Fig. 4-17 MR data at $T = 65.5\, \text{K}$ of (a) 100 nm and (b) 200 nm thick 2 $\mu\text{m}$ Fe wires.

<table>
<thead>
<tr>
<th>Thickness (nm)</th>
<th>25</th>
<th>50</th>
<th>100</th>
<th>200</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\rho_0(1.5, \text{K})$ ($\mu\Omega\text{cm}$)</td>
<td>1.2</td>
<td>1.48</td>
<td>0.71</td>
<td>0.45</td>
</tr>
<tr>
<td>$\rho_0(-65, \text{K})$ ($\mu\Omega\text{cm}$)</td>
<td>1.84</td>
<td>2.14</td>
<td>2.1</td>
<td>0.98</td>
</tr>
<tr>
<td>RRR</td>
<td>11.8</td>
<td>9.5</td>
<td>25.6</td>
<td>31.7</td>
</tr>
<tr>
<td>Domain Size $d$ ($\mu\text{m}$)</td>
<td>0.64</td>
<td>0.45</td>
<td>0.45</td>
<td>0.58</td>
</tr>
<tr>
<td>$\rho_d/\rho_0(H=0, , T=1.5, \text{K})$</td>
<td>-0.5 %</td>
<td>-0.75 %</td>
<td>-1.2 %</td>
<td>-0.6 %</td>
</tr>
<tr>
<td>$\rho_d/\rho_0(H=0, , T=65, \text{K})$</td>
<td>-0.06 %</td>
<td>-0.1 %</td>
<td>-0.1 %</td>
<td>0.0 %</td>
</tr>
<tr>
<td>$r=\rho_d\cdot d$ (1.5 K) ($\Omega\text{m}^2$)</td>
<td>$-3.8 \cdot 10^{-17}$</td>
<td>$-4.9 \cdot 10^{-17}$</td>
<td>$-3.8 \cdot 10^{-17}$</td>
<td>$-1.6 \cdot 10^{-17}$</td>
</tr>
</tbody>
</table>

Table 4-1 Characteristic data for 2 $\mu\text{m}$ linewidth wires as a function of thickness.
given by \( r = \rho_d d \), where \( d \) is the domain size, is largest in the thinner films (\( \leq 100 \) nm).

There are a few models of DW scattering which predict enhancements in the conductivity in the presence walls. One is that of Tatara and Fukuyama based in weak localization phenomena [4-7]. They introduce a wall decoherence time to parametrize this effect \( \tau_w = \tau/[n_w/(6\lambda k_f^2)\varepsilon_f/\Delta^2] \). Here \( \tau \) is the momentum relaxation time, \( n_w \), the domain wall density, \( k_f \) the Fermi vector, \( \lambda \) the domain wall thickness, and \( \varepsilon_f/\Delta \) the ratio of the Fermi energy to the exchange splitting of the band. With commonly used parameters for s electrons in Fe, \( \varepsilon_f/\Delta \sim 500 \), \( k_f \sim 1.7 \) \( \text{Å} \), \( \lambda \sim 300 \) \( \text{Å} \) and with \( n_w = 2.5 \) \( \mu \text{m}^{-1} \) we estimate \( \tau_w \sim 60\tau \). Essential to observing such an effect is the absence of other decoherence mechanism, such as inelastic scattering. Equating \( \tau_w = \tau_\text{in} \) gives an upper temperature limit for the presence of weak localization phenomena. From the residual resistance \( \tau = 2.8 \times 10^{-14} \) s and with \( \rho_\text{in} = \alpha T^2 \) (\( \alpha = 3 \times 10^{-4} \)) we find \( T_{\text{max}} = 7 \) K. The suppression of weak localization due to DWs does not explain our observation of enhanced conductivity up to \( \sim 80 \) K.

Our results suggest a novel mechanism by which domain walls may increase conductivity in thin films. The interplay between orbital effects in the internal magnetic fields near DWs and surface scattering may be at the origin of the anomalous negative wall contribution. 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
Fig. 4-18 Effects of internal fields and surface scattering on the trajectory of charge
to deflect charge from the films interfaces and hence reduce resistivity (Fig. 4-18). It appears that the magnetic structure with approximately a mean free path (∼ 200 nm at 1.5 K in a 100 nm thick film) is important to the observed phenomena. Increasing the film thickness acts to reduce the importance of the surface scattering and hence this effect. At higher temperature, the mean free path gets shorter, and surfaces produce less of the scattering and reduce this effect.

4.4 Domain wall scattering in hcp Co

4.4.1 magnetic characterization

Co layers of thickness of 55 nm, 70 nm, 145 nm, and 185 nm have been prepared. Similar transport structures as Fe have also been patterned. A residual resistivity of 0.16 μΩcm and the residual resistivity ratio of 19 for a 185-nm-thick Co wire confirm the high crystalline quality of the films. These films have strong uniaxial anisotropies with the magnetic easy axis perpendicular to the film plane. The competition of magnetic energies leads to stripe domains in which the domain size now depends on the sample thickness [Fig. 4-19]. Thicker films have larger domains. The domain configurations also depend on the sample magnetic history. Figure 4-20 shows MFM images of a 70-nm-thick 5-μm-wide Co wire in zero magnetic field. Images are taken 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. As seen in Fig. 4-20, an in-plane applied field can be employed to align DWs in stripes. Figure 4-20 (b) and (c) show that DWs can be oriented parallel or
Fig. 4-19 MFM images of stripe domains in 55 nm, 70 nm, 145 nm, and 185 nm Co films.
Fig. 4-20 MFM images in zero applied field of a 5-\textmu m-linewidth 70-nm-thick Co wire after (a) perpendicular, (b) transverse, and (c) longitudinal magnetic saturation.

The model
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 (as shown in the drawing in Fig. 4-20).

Modeling of the film micromagnetic structure is essential to understand the shows the orientation of stripe and flux closure caps with respect to the current for (b) CPW and (c) CIW geometries. MR results. An important parameter for stripe domain materials is the ratio of anisotropy to demagnetization energy, known as the quality factor Q, given by $Q = \frac{K}{2\pi M_s^2}$. Since hcp Co has an intermediate Q value ($Q = 0.35$), numerical modeling of film micromagnetic structure is necessary to determine equilibrium domain configurations. It has been shown numerically that in hcp Co DW’s branch, being Bloch–like in the film center and forming flux closure caps at the top and bottom surface of the film to reduce the magnetostatic energy [4-11].

The magnetic structure of films of the thickness studied has been computed in zero field with LLG Micromagnetic Simulator [4-12]. The basics of micromagnetic simulation will be discussed in Chapter 5. The equilibrium state is found from the minimization of the system’s free energy. As seen in Fig. 4-21 (a), such calculations produce domain widths, which are in good 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. The inset of Fig. 4-21 (b) shows a part of the simulated magnetic cross section of a 70-nm-thick Co elements (with overall dimensions of $1500\times500\times70$ nm), where the arrows indicate the magnetiza-
Fig. 4-21 (a) Domain size versus film thickness; experimental (solid circles) and calculated values (solid squares). (b) The calculated in-plane magnetization volume (solid squares) and the magnitude of the MR $R_{P,\text{meas}} - R_{P,0}$ in the perpendicular geometry (solid circles) as a function of wire thickness. Inset: Calculated magnetic domain cross section of a 70-nm-thick Co element showing out-of-plane magnetized stripe domains and in-plane magnetized flux closure caps.
tion direction of the stripe and flux closure caps. Flux closure caps constitute approximately 25% of the total wire volume, which is also an approximate measure of the in-plane magnetization volume. For all Co wire thickness investigated the closure cap volumes (in-plane magnetization) were calculated as shown on the left-hand axis in Fig. 4-21 (b). By increasing the wire thickness from 55 to 185 nm the in-plane magnetization volume decrease from 33% to 17%.

4.4.2 Transport Properties

The general features of the MR of these materials are similar to those seen in Fe microstructures. Figure 4-22 shows MR results for the three different field orientation at 280 K, 85 K, and 1.5 K. The low field MR is negative for fields applied along the magnetic easy axis (i.e. perpendicular to the field plane), and positive for in-plane applied fields (transverse and longitudinal geometry). Hysteresis is also evident, particularly in the perpendicular case, which correlates well with magnetization hysteresis loops (Fig. 4-23). At 280 K, above the saturation field of ~ 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 plane and parallel to the current (M // 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 M and J, but also on the direction that M makes with respect to the crystal axes, with the resistivity smallest for M ⊥ J and parallel to the [0001] direction. At low temperature (85 K and 1.5 K) the resistivity is largest above the
Fig. 4-22 MR data of a 5-μm-linewidth 55-nm-thick Co wire in the perpendicular, transverse, and longitudinal field geometries at (a) room temperature, (b) 85 K, and (c) 1.5 K.
Fig. 4-23 Magnetization hysteresis loops measured with a superconducting quantum interference device magnetometer of a 55-nm-thick Co sample at 300 K for applied fields in plane (dashed line) and perpendicular to the film plane (solid line).
saturation field in the transverse geometry. As in the case of Fe, and for the reasons already discussed, the in-plane resistivity anisotropy change sign with temperature.

The resistivity anisotropy is again important in the interpretation of the low field MR because the magnetization in zero applied field has components along all three dimensions. For example, in the CPW geometry (as illustrated in Fig.4-20), the magnetization of the stripe domain are out-of-plane and perpendicular to the current, the magnetization of the flux closures caps are in plane and parallel to the current, and the magnetization of the Bloch wall rotates through an orientation in-plane and perpendicular to current. Thus a saturation field will both erase DWs and reorient the magnetization with respect to the current and crystal.

This contribution can be estimated again with an effective medium model of the resistivity. In the limit in which the electron mean free path is smaller than the domain size (at room temperature we estimate the electron mean free path to be 30 nm), and since the resistivity anisotropy is small, the resistivity can be written as a weighted average of the resistivities, to first order in the resistivity anisotropy, \( \varepsilon_L = R_{L,0} - R_{P,0} \), and \( \varepsilon_T = R_{T,0} - R_{P,0} \). Starting from the maze configuration (Fig. 4-20 a) the perpendicular MR is

\[
R_{P,\text{meas}} - R_{P,0} = \gamma[(R_{L,0} + R_{T,0})/2 - R_{P,0}]
\]

(4-9)

where \( \gamma \) is the volume of in-plane magnetized closure caps. Here \( R_{L,T,P,0} \) are the MR extrapolated from high field to \( H = 0 \) 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. For the MR measurements shown in Fig. 4-22 (a) 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 magnitude of the perpendicular MR is thickness dependent because the volume of the in-plane magnetization material depends on sample thickness (Fig. 4-21 b). Fig. 4-21 b) shows that the perpendicular MR generally increases with increasing in-plane magnetized volume fractions.

The measured difference between CPW and CIW resistivities (i.e. $R_{l,\text{meas}} - R_{l,\text{meas}}$) in Fig. 4-22 a) is given in terms of the resistivity anisotropy as

$$R_{l,\text{meas}} - R_{l,\text{meas}} = \gamma (R_{l,0} - R_{T,0})$$

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 predominate MR effects observed in this material are explicable by film microstructure 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.

Temperature-dependent resistivity measurements for CPW and CIW geometries show more interesting behavior, which is not explicable simply in terms of ferromagnetic resistivity anisotropy. With decreasing temperature the in-plane resistivity anisotropy change sign ($T < 65 \text{ k}$) (Fig. 4-22), as we discussed for the Fe microstructures. At $T = 85 \text{ K}$, this in-plane resistivity anisotropy is nearly zero ($R_{l,0} = R_{T,0}$). Thus the difference between CPW and CIW resistivities due to the in-plane
resistivity should also approach zero. But a small measured difference in CPW and CIW resistivity, \( \Delta \rho = R_{\text{odm}} - R_{\text{t.m}} = 9.4 \times 10^{-4} \), is observed. And \( \Delta \rho \) is always positive with temperature, and almost a constant, while the resistivity anisotropy changes sign and become larger in magnitude with decreasing temperature [Fig. 4-24]. This implies that the difference between CPW and CIW resistivity is not simply due to resistivity anisotropy.

Temperature dependent MR measurements in the longitudinal and transverse field geometries are consistent with a DW scattering mechanism proposed by Levy and Zhang [4-5]. As in GMR, this model predicts a greater DW-MR for CPW than for the CIW geometry, as we observed in experiments. If we assume that the difference is due to this novel mechanism it is possible to estimate the DW interface resistance. Levy and Zhang calculated a ratio of CPW to CIW MRs of:

\[
\frac{R_{\text{CPW}}}{R_{\text{CIW}}} = 3 + \frac{10 \sqrt{\rho_0^+ \rho_0^-}}{\rho_0^+ + \rho_0^-}.
\]

(4-11)

where \( \rho_0^+ \) and \( \rho_0^- \) are the spin up and spin down resistivities. Taking the spin asymmetry for Co [4-11] to be \( \rho_0^+/\rho_0^- = 3 \), Equation (4-11) gives \( R_{\text{CPW}} / R_{\text{CIW}} = 7 \). And from the measured difference, \( \Delta \rho \), both \( R_{\text{CPW}} \) and \( R_{\text{CIW}} \) can be estimated. We find that \( R_{\text{CPW}} = 1.1 \times 10^{-3} \). This gives an MR just due to the DW material of \( \Delta \rho_{\text{wall}}/\rho_0 = \frac{d}{\delta} R_{\text{CPW}} = 0.5\% \), where \( d \) is the domain size (66 nm), \( \delta \) is the wall width (\( \sim 15 \) nm) and \( \rho_0 \) is the film resistivity (at \( T_{\text{comp}} \), \( \rho_0 = 0.9 \) \( \mu \Omega \text{cm} \)). Further, the DW interface resistance for the CPW geometry is given by \( \tau = \Delta \rho_{\text{wall}} \delta = 7 \times 10^{-19} \Omega \text{m}^2 \) at \( T_{\text{comp}} \). At
Fig. 4-24 Temperature dependence of difference between CPW and CIW resistivities, \( \Delta \), and \( R_{L,0} - R_{T,0} \) of a 55-nm-thick Co wire.

Table 4-2: Characteristic data for 5 \( \mu \)m linewidth Co wires of 55 nm, 70 nm, 145 nm, and 185 nm thickness.

<table>
<thead>
<tr>
<th>Thickness (nm)</th>
<th>55</th>
<th>70</th>
<th>145</th>
<th>185</th>
</tr>
</thead>
<tbody>
<tr>
<td>d (nm)</td>
<td>66</td>
<td>80</td>
<td>116</td>
<td>135</td>
</tr>
<tr>
<td>( \rho_0 (1.5 \text{ K}) ) (( \mu \Omega \text{cm} ))</td>
<td>0.63</td>
<td>0.26</td>
<td>0.23</td>
<td>0.16</td>
</tr>
<tr>
<td>( \rho_0 (T_{\text{comp}}) ) (( \mu \Omega \text{cm} ))</td>
<td>0.92</td>
<td>0.68</td>
<td>0.58</td>
<td>0.3</td>
</tr>
<tr>
<td>( \rho_0 (RT) ) (( \mu \Omega \text{cm} ))</td>
<td>3.83</td>
<td>3.04</td>
<td>3.31</td>
<td>3.04</td>
</tr>
<tr>
<td>( \Delta u (T_{\text{comp}}) )</td>
<td>0.94 ( \cdot 10^{-3} )</td>
<td>0.75 ( \cdot 10^{-3} )</td>
<td>1.3 ( \cdot 10^{-3} )</td>
<td>1.4 ( \cdot 10^{-3} )</td>
</tr>
<tr>
<td>( r(T_{\text{comp}}) ) (( \Omega \text{m}^2 ))</td>
<td>5.7 ( \cdot 10^{-19} )</td>
<td>4.1 ( \cdot 10^{-19} )</td>
<td>8.7 ( \cdot 10^{-19} )</td>
<td>5.7 ( \cdot 10^{-19} )</td>
</tr>
</tbody>
</table>
1.5 K, taking account of the resistivity anisotropy with Equation (4-10), \( r \) is \( 2 \times 10^{-18} \) \( \Omega \) m\(^2\). Table 4-2 summarizes the MR measurements at the compensation temperature and these estimations for different wire thickness. For the films studied the average interface resistance is \( 6 \pm 2 \times 10^{19} \) \( \Omega \) m\(^2\). For comparison, this is a factor of 10 to 100 smaller than Co/Cu interface resistance in GMR multilayers with current perpendicular to the plane of the layers [4-13].

4.5 Domain wall scattering in FePt

Most theories predict that narrower DWs will produce larger scattering. Thus materials with very high uniaxial anisotropy are ideal for isolating the intrinsic domain wall scattering effect, since higher anistropy leads to narrower walls. It also provides further control of magnetic domain configurations in restricted geometries. Next, we present results on chemically ordered (001) \( L1_0 \) FePt epitaxial thin films with perpendicularly magnetized stripe domains. These thin films have among the highest known magnetic anisotropy energy (\( K \sim 10^8 \) erg/cm\(^3\)) of any ferromagnetic material.

4.5.1 Experiment

Epitaxial (001) oriented \( L1_0 \) Fe\(_{1-x}\)Pt\(_x\) (\( x \sim 0.5 \)) thin films (100 nm thick with thin Pt seed and cap layers) were grown by MBE in ultrahigh vacuum on (001) MgO substrates as described in [4-14]. The substrate temperature was varied between 150 and 500 °C. X-ray diffraction analysis was used to determine the degree of chemical order \( (S/S_{\text{max}}) \) and film composition was determined by Rutherford back scattering.
(RBS) analysis [4-14]. The room temperature magnetic properties have been measured using both torque and vibrating sample magnetometry.

The films were patterned using optical lithography and ion milling to produce 20 μm linewidth wires with contacts for measurement of both longitudinal and transverse (Hall) resistivities (Fig. 4-25). MFM imaging and magnetotransport measurements were performed. The applied field was oriented perpendicular to the film plane (and current) as well as in the film plane and parallel to the current, denoted the longitudinal field geometry. A 4 probe ac (~10 Hz) resistance bridge and low bias currents (100 to 200 μA) were employed.

4.5.2 Results

Table 4-3 summarizes the properties of the films studied. With increasing substrate temperature there is a greater degree of chemical order and a higher uniaxial anisotropy constant. Torque measurements on samples 1079 and 1080, (and to a lesser extent on 1075) also indicate higher order components to the anisotropy, and that these films may be inhomogeneously ordered [4-15].

Figure 4-26 shows MFM images at room temperature of films 1075 and 1080. Prior to imaging, the films were demagnetized with a field applied perpendicular to the film plane. The light and dark contrast in images is associated with magnetization parallel or antiparallel to the film normal. The average domain size is larger in the higher anisotropy film (1075, ~ 200 nm). Domains form due to a competition between magnetic energies: the exchange, dipolar (magnetostatic) and
Fig. 4-25 Optical micrograph of 20 μm FePt transport structure.

TABLE 4-3. Structural, magnetic, and transport characteristics of the films studied. \( \delta \) is an estimate of the domain wall width \([\pi (A/K_v)^{1/2}, \text{with } A = 10^6 \text{ erg/cm}], d \) is the average domain size, \( \rho \) is resistivity, and RRR is the residual resistivity ratio. Other symbols are as defined in the text.

<table>
<thead>
<tr>
<th>( T_s ) (C)</th>
<th>( x )</th>
<th>( S/S_{\infty} )</th>
<th>( M_s ) (emu/cm(^2))</th>
<th>( K_v ) (10(^6) erg/cm(^2))</th>
<th>( \delta ) (nm)</th>
<th>( d ) (nm)</th>
<th>( \rho (\mu \Omega \text{cm}) ) (1.7 K)</th>
<th>RRR</th>
<th>Domain MR (1.7 K)</th>
<th>( \rho_{\infty}/\rho_{\infty} ) (1.7 K)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1075</td>
<td>500</td>
<td>0.49</td>
<td>0.80</td>
<td>834</td>
<td>4.30</td>
<td>4.4</td>
<td>210</td>
<td>7.2</td>
<td>3.0</td>
<td>2.8\times10^-1</td>
</tr>
<tr>
<td>1079</td>
<td>250</td>
<td>0.56</td>
<td>0.63</td>
<td>745</td>
<td>-0.31</td>
<td>18</td>
<td>165</td>
<td>18.5</td>
<td>2.2</td>
<td>1.2\times10^-1</td>
</tr>
<tr>
<td>1080</td>
<td>150</td>
<td>0.51</td>
<td>0.47</td>
<td>837</td>
<td>-0.28</td>
<td>19</td>
<td>90</td>
<td>26.6</td>
<td>1.8</td>
<td>1.1\times10^-1</td>
</tr>
</tbody>
</table>
Fig. 4-26  MFM images in zero applied field of Fe$_{1-x}$Pt$_x$ (a) 1075, a well-ordered, high anisotropy film and (b) 1080, lower growth temperature and anisotropy.
anisotropy energies. For film 1075 we estimate the ratio of magnetcocrystalline to magnetostatic energy density, \( Q = K/2\pi M^2 = 10 \). For such a large \( Q \), stripe domains which intersect the surface with \( \mathbf{M} \) perpendicular to the film surface are energetically favored. Qualitatively larger anisotropy leads to a greater DW energy and hence larger domains, as observed.

Figure 4-27 a) shows MR measurements with the field oriented perpendicular (solid lines) to the film plane at 280 K and 1.7 K. In the low field region MR hysteresis is observed which correlates well with magnetic hysteresis measurements. At low temperature, the high field MR is positive and quadratic with field. This we associate with the ordinary (Lorentz) MR. At 280 K the high field MR is negative and decreases nearly linearly out to the largest fields that we are able to apply (10 T). Such a negative MR is typically associated with the magnetic field suppression of spin-disorder scattering [4-10,16].

Measurements in the longitudinal field geometry (with the field at 5 degrees to the film plane) are indicated by dashed lines in figure 4-27. Resistivity anisotropy is noticeable at both high and low temperatures. At low temperature (1.7 K) differences are observed particularly when magnetic domains are present, between –8 T and 8 T. At 280 K, the resistivity in the longitudinal geometry is generally larger than that in the perpendicular geometry. This is again associated with magnetic domain structure, which we discuss below.

Fig. 4-27 (b) shows the transverse or Hall resistivity as a function of field at 1.7 K. These characteristics are again hysteretic, reflecting the magnetic hysteresis.
As commonly observed in ferromagnetic materials there is an ordinary linear high field component and an extraordinary component proportional to the sample magnetization [4-15]. The extraordinary component is associated with the spin-orbit interaction, which leads to both asymmetric scattering (skew scattering) and a side-jump mechanism for the Hall effect. This latter contribution is predicted to scale as $\rho_{xx}^{-2}$ [4-16]. The inset of Fig. 4-27 b) shows a log-log plot of $\rho_{xy}$ versus $\rho_{xx}$. We observe $\rho_{xy} \sim \rho_{xx}^{-2}$, consistent with the side-jump mechanism.

Figure 4-28 shows similar magnetotransport measurements on a lower anisotropy film (1080) with greater chemical disorder. At the lowest temperature the resistivity is nearly field independent, and indicates a suppression of the Lorentz MR with increasing film disorder. At higher temperature a negative linear high field MR is observed. Figure 4-28 (b) shows that the extraordinary Hall angle is greatly enhanced with respect to that of 1075, consistent with the side jump mechanism $(\rho_{xy}/\rho_{xx} - \rho_{xx})$.

4.5.3 Discussion

These magnetotransport results illustrate that domain structure has a significant effect on film resistivity in such materials. For instance, in film 1075, the resistivity is enhanced by 0.2 % to 0.3 % at low fields and temperatures due to the presence of magnetic domains. We denote this enhancement the domain MR. At 280 K a smaller enhancement in the resistivity is observed near zero field in the perpendicular MR. The larger resistivity in the longitudinal geometry may be associ-
Fig. 4-27  a) MR data of a 20 μm linewidth wire of FePt 1075 at 1.7 K and 280 K. The solid line is with the applied field oriented perpendicular to the film plane and the dashed line is for the field oriented 5 degrees from the film plane and parallel to the current. The field is purposely misaligned from the plane by this angle so that the sample is in a well-defined (single domain) magnetic state at high field. b) The Hall angle, \( \rho_{xy}/\rho_{xx} \), versus perpendicularly applied field at 1.7 K. The inset in b) shows a log-log plot \( \rho_{xy} \) versus \( \rho_{xx} \).
Fig. 4-28 a) MR data of a 20-μm-linewidth wire of FePt 1080 at 1.7 K and 280 K in a perpendicular applied field. b) The Hall angle versus perpendicularly applied field at 1.7 K. The inset shows a log-log plot of $\rho_{xy}$ versus $\rho_{xx}$. 
ated with the orientation of the domains. MFM imaging shows that after longitudinal measurements at 280 K, DWs tend to align perpendicular to the current, instead of in the maze like pattern seen in Fig. 4-26. Low field enhancements in resistivity, somewhat smaller in magnitude (0.1 \%) are also observed in film 1080.

An important question is the physical mechanism of this resistivity enhancement—whether it is due to an *intrinsic* DW scattering contribution to the resistivity or "domain" effects. A number of domain effects could enhance the resistivity. Ferromagnetic resistivity anisotropy is a mechanism by which a multidomain sample may have a higher resistivity than that of a single domain sample. Berger has also discussed a domain mechanism which can enhance resistivity based on the Hall effect, and the resistivity is predicted to be increased of order \((\rho_{xy}/\rho_{xx})^2\).

In the high anisotropy film (1075), these domain effects appear to be of insufficient magnitude to explain the observed increase in resistivity. For the Hall effect mechanism the observed increase of 0.3 \% is more than one order of magnitude larger than \((\rho_{xy}/\rho_{xx})^2 \sim 0.01\% at 1.7 K. Also, while the Hall angle decreases as the temperature is reduced, the domain MR increases (Fig. 4-29). Further, the ferromagnetic resistivity anisotropy is small and due to the large uniaxial anisotropy domain magnetization is mainly perpendicular to the film plane and hence current. Film 1080 has a lower anisotropy, a smaller domain size and hence a larger density of DWs, yet the magnitude of the domain MR is reduced, \(0.1\ \%\). In this film, \((\rho_{xy}/\rho_{xx})^2\) is also same order of magnitude and the Hall mechanism may be rela-
vant. Thus in the high anisotropy film an intrinsic spin dependent DW scattering contribution to the resistivity may be at the origin low field enhancement of the resistivity.

To summarize this chapter, I have studied the DW scattering on conduction electrons in bcc Fe, hcp Co, and L1₀ FePt. AMR and Lorentz MR are predominant in Fe and Co structures, while in FePt evidence for an intrinsic spin-dependent DW scattering contribution to the resistivity has been found.
Chapter 5 Magnetization Reversal In Fe Particles

5.1 Introduction

Micron scale patterned thin film ferromagnetic elements are a topic of great interest [5-1, 2, 3] with important applications to magnetic information storage technology [5-4, 5]. Magnetic behavior such as, hysteresis, magnetization reversal, and domain configurations, change significantly from that in extended thin films due to the presence of lithographically defined boundaries [5-1, 2, 3]. For in-plane magnetized films, these boundaries increase the importance of dipolar interactions or shape anisotropy, as a normal component to the magnetization at the boundary leads to magnetic charges [5-6]. The magnetic properties are in general a consequence of the interplay between magnetostatic, exchange, magneto-crystalline and Zeeman energies. Thermal fluctuations may also play an important role in the magnetization reversal, and become particularly important in nanometer scale ferromagnetic structures. While at micron scales sizes individual elements have only a small number of domains [5-1, 2, 3], the magnetic behavior is still in general quite complex, and challenging to model using either analytic or numerical micromagnetic techniques [5-1, 7, 8]. An understanding of magnetic phenomena in micron scale elements is nonetheless essential to applications of these materials in magnetoelectronics and experiment can also guide micromagnetic modeling. As shown in this chapter, micron scale materials with well-characterized magnetic interactions can provide important insight into the
control of magnetic characteristics by shape and the effects of magnetic topology on magnetization reversal.

I have examined the physical consequences of element shape and anisotropy on magnetic phenomena in this size range by studying model thin film microstructures. Epitaxial growth and advanced microfabrication techniques have been used to produce elements with controlled magnetocrystalline anisotropy and shape. Magnetization reversal and domain configurations are studied in individual magnetic elements by direct imaging of domains in applied magnetic fields with a magnetic force microscope (MFM). Magnetic hysteresis is studied in arrays of nominally identical elements using longitudinal Kerr hysteresis loop measurements (MOKE). This combination enables an understanding of the connection between local magnetic domain configurations within elements and magnetic hysteresis of an array of such elements (such as, the coercivity, remanent magnetization and switching characteristics). Micromagnetic simulation has also been done on these elements. Good agreement between experimental results and micromagnetic calculations allows a precise description of the magnetization processes in these systems.

5.2 Fabrication and characterization

(110) oriented 50 nm thick Fe films were grown on a-axis (11-20) sapphire substrates using UHV e-beam evaporation techniques. These Fe films have been patterned into particle arrays with rectangular, triangular, and needle-shaped ends using electron beam lithography and ion-milling [Fig. 2-9]. Details of the fabrication
processes are described in chapter 2. The magnetic properties and the magnetization reversal process of arrays of particles were studied via magneto-optic Kerr effect (MOKE) hysteresis loops and individual particles were studied with a MFM, which incorporated an in-situ electromagnet. The later enabled fields up to 1500 Oe to be applied during imaging experiments. Figure 2-9 shows scanning electron microscope (SEM) images of these particles with rectangular, triangular, and needle-shaped ends. The rectangular particles are 2 μm or 0.5 μm wide and have a length-to-width ratio of 3:1, while the other two particle types have two additional triangular or needle-like ends. Thus, the overall length of the needle is nine times its width. All particles are separated by 4 times their width and 3 times their length, so that dipolar interaction between neighboring particles within the array is negligible. Indeed, the dipolar fields are estimated to be less than 10 Oe from micromagnetic calculations, assuming the particles are saturated along their long axes [5-9]. The particles have been oriented either with their long axis perpendicular to the in-plane magnetocrystalline easy [001] axis or with their long axis parallel to the easy [001] axis. In the former case a competition between magnetic energies results, whereas in the latter case the magnetocrystalline anisotropy energy reinforces the element shape anisotropy.

5.3 Results and discussion

5.3.1 Magnetic easy axis perpendicular to the long particle axis

First we consider the case in which the particle long axis is perpendicular to the [001] easy axis, for which a competition between exchange, magnetocrystalline
and magnetostatic energy occurs. Figure 5-1 shows the hysteresis loops of 2 μm wide rectangular and needle-like end particles in longitudinal and transverse field geometries. H-longitudinal refers to the applied magnetic field parallel to the long axis of the particle while H-transverse refers to the field perpendicular to the long axis (i.e., H // [001]).

In the longitudinal case [Fig. 5-1 a)] the hysteresis loops of rectangular and needle-like end particles both exhibit a small remanence (≈ 0.09 M_s) at H=0. For the rectangular particles the saturation field H_s and the nucleation field H_N are H_s=400 Oe and H_N=300 Oe, respectively. For needle-like particles H_s is 300 Oe and H_N is 200 Oe.

In the transverse case [Fig. 5-1 b)] the rectangular particles show nearly full remanence, a nucleation field of approximately H_N=0 Oe, and magnetic saturation field of H_s=400 Oe. In contrast, the needle-shaped particles show a smaller remanence and a nucleation field of H_N=100 Oe and a saturation field of H_s=700 Oe.

Fig. 5-2 shows the hysteresis loops of 0.5 μm wide rectangular and needle-like end particles in longitudinal and transverse geometry. In the transverse case [Fig. 5-2 a)] the hysteresis loops of rectangular and needle-like end particles both exhibit almost a fully demagnetized state at H=0. For rectangular particles the saturation field H_s and the nucleation field H_N are determined to be H_s=1700 Gauss and H_N=400 Gauss. For needle-like particles H_s is 2000 Gauss and H_N is 750 Gauss. The particles with needle-shaped ends have higher nucleation and saturation fields.

In the longitudinal case [Fig. 5-2 b)] the rectangular particles show a fully demagnetized state at zero applied field, a nucleation field of only H_N=250 Gauss, and
the magnetic saturation occurs at only $H_s=450$ Gauss. In contrast, the needle-like end particles show a square hysteresis loop with full remanence and a saturation field of only $H_s=250$ Gauss.

MFM measurements of these Fe particles has been performed in order to correlate these magnetic hysteresis loops with the corresponding magnetization reversal processes and magnetic domain patterns of individual particles. Also, the influence of the particles shape and the magnetic history on the nucleation process of magnetic domains and on the magnetic domain configurations at $H=0$ has been studied. The NiFe coated MFM tips used are vertically magnetized.

Figure 5-3 presents MFM images of the magnetic domain configurations of 2 μm rectangular and needle-shaped particles at selected magnetic fields in the longitudinal field geometry. In the case of the rectangular particle [Fig. 5-3 a) to e)], domains are visible first at the element corners at an applied field of approximately 420 Oe. Figure 4(a) reveals a magnetization pattern with domain walls at the particle ends and edge domains. In this particular element, both edge domains are larger on the top edge of the particle, which implies that the magnetizations at the left and right edges are in opposite directions. This is sometimes called a “C” state. On reduction of the field [Fig. 5-3 b)] a vortex is formed and domains move towards the center of the particles. The arrows in Fig. 5-3 e) indicate the in-plane magnetization direction of the stripe domains with the magnetization parallel or antiparallel to the easy magnetocrystalline axis and the flux closure domains where the magnetization is directed perpendicular to the easy axis. The MFM image of the demagnetized state at
Fig. 5-1 Longitudinal MOKE hysteresis loop measurements on 2 μm wide rectangular and needle-shaped (110) Fe particles with the applied magnetic field a) parallel (H-longitudinal) and b) perpendicular (H-transverse) to the long axis of the particles.
Fig. 5-2 Longitudinal MOKE hysteresis loop measurements on 0.5 \( \mu \)m wide rectangular and needle-shaped (110) Fe particles with the applied magnetic field perpendicular (H-transverse) and parallel (H-longitudinal) to the long axis of the particles.
H=0 [Fig. 5-3 e]) is consistent with the small remanence seen in MOKE hysteresis loops [see Fig. 5-1 a]). In contrast, for the needle-shaped particles the nucleation of magnetic domains occurs at lower magnetic field (≈ 195 Oe) and closer to the particle center. In the demagnetized state at H=0 Oe the width, d, of the stripe domains in the center of the needle-shaped particle [see Fig. 5-3 j) right side] is significantly smaller than that observed in the rectangular particle [Fig. 5-3 e]). The average value of the domain width near the center of the elements is determined to be $d_n=0.7 \, \mu\text{m}$ and $d_k=1.0 \, \mu\text{m}$ for the needle-shaped and the rectangular particle, respectively.

The MFM images in Fig. 5-4 and Fig. 5-5 give an overview of the nucleation and motion of magnetic domains in the presence of a transverse applied magnetic field. For both shapes, images are shown on decreasing the magnetic field from the magnetic saturation field. Figure 5-4 shows the evolution of magnetic domains for a rectangular particle in the transverse field geometry. Starting from a positive magnetic field, domains nucleate close to the particle corners. But, as discussed in detail in a following section, the location of the nucleation also depends strongly on the magnetic history. The hysteresis loop in transverse geometry in Fig. 5-1 b) indicates that the particle is almost fully saturated above 400 Oe, while the MFM images in Fig. 5-4 a) and j) show very small domains along the long edges of the particle. For increasing negative fields, where two domain walls approach each other [see Fig. 5-4 g) to j)], a field much larger than the coercive field is necessary to annihilate the domain walls and reversed magnetic domains near the center of the particle. For saturation fields higher than approximately 500 Oe the nucleation occurs at fields of only 40 to 60 Oe [
see Fig. 5-4 c) and d)]. This is not consistent with magnetic hysteresis measurements, as the hysteresis loop in Fig. 5-1 b) indicates almost full remanence, whereas the MFM image in Fig. 5-4 e) shows a more or less demagnetized state. The stray field of the MFM tip may be influencing the nucleation process. Likely the stray field has initiated the nucleation of domains and, therefore shifts the nucleation field to higher values in comparison with that determined via MOKE measurements. The domain structure observed in zero field does not appear to be affected by the MFM tip, as the images are both stable in time and observed on a number of different particles after the same magnetic preparation. Also note that in this geometry the stripe domain width is \( d_s = 2.8 \ \mu m \) in zero field and is larger than that in the longitudinal field geometry.

For the needle-shaped particle (see Fig. 5-5) the nucleation of domains takes place first at the outermost part of the needle, which is in clear contrast to the longitudinal field geometry, and occurs in a field range of 100 to 200 Oe. Like the rectangular particles, the average domain width of \( d_N = 1.9 \ \mu m \) in the center of the particle is clearly increased in comparison with the longitudinal geometry.

The interplay between exchange, Zeeman, shape and magnetocrystalline anisotropy energies determines the micromagnetic structure and magnetization reversal process. With decreasing particle size, the demagnetizing field arising from the magnetic charges formed at particle edges becomes important. This introduces a shape anisotropy with the easy axis parallel to the long axis of the particle, and perpendicular to the magnetocrystalline easy axis. As showed in Chapter 1, a stripe domain configuration minimizes the free energy in such a system. Due to the small
Fig. 5-3 Field dependent MFM images on 2 µm wide a)-e) rectangular and f)-j) needle-shaped (110) Fe particles in longitudinal field geometry.
Fig. 5-4 Field dependent MFM images of the magnetic domain configuration in a 2 μm wide rectangular particle in transverse field geometry.
Fig. 5-5 Field dependent MFM images of the magnetic domain configuration in a needle-shaped particle in transverse field geometry.
ratio of magnetocrystalline anisotropy to demagnetization energy \( Q = (K_{ij} + K_w)/2\pi M^2 \) \( \approx 0.05 \), stripe domains in conjunction with flux closure domains at \( H=0 \) are energetically favored, such as observed in our microfabricated transport Fe wires [see section 4.3.2].

In the transverse case, both for rectangular and needle-shaped particles, domains nucleate near the particle ends on reducing the applied field from its saturation value. The demagnetizing fields generated when the magnetization is rotated away from the long particle axis are much larger at the outermost end of needle-shaped particles than at the corners of the rectangular particles. Therefore, higher nucleation and saturation fields are observed for needle-like end particles. After nucleation of magnetic domains at the ends of the particle, the domains move towards the particle center forming a flux closure multidomain state at \( H=0 \).

For the longitudinal field geometry both the shape anisotropy and Zeeman energy favor a magnetization parallel to the applied field. The enhanced shape anisotropy at the outermost part of the needles favors an alignment of the magnetization parallel to the particles long axis. Hence, the nucleation and saturation fields for the needle-like particle are slightly lower than these observed for the rectangular particle.

The shape anisotropy has a more remarkably impact on the location where the magnetic domains nucleate first. In case of the needle-shaped particle the enhanced shape anisotropy near the long tail of the needle suppresses a formation of domains at the outmost part of the needle. In narrower (i.e. smaller width) particles, the shape
anisotropy can overwhelm the magnetocrystalline anisotropy and suppress the nucleation of domains completely, resulting in a fully saturated state at $H=0$. Figure 5-6 shows domain structures of $0.5 \ \mu \text{m}$ wide particles at $H=0$. After longitudinal saturation, the needle-like particle shows a single-domain state [Fig. 5-6 c)], which is in agreement with the corresponding MOKE measurement [Fig. 5-2 a)]. In contrast, after transverse saturation it shows a multidomain state. The rectangular particle shows multidomain configuration in both cases, but it has two central domains after longitudinal saturation and only one after transverse saturation.

In the transverse field geometry the topology of the magnetization during reversal has an important influence on hysteresis. Figure 5-7 shows normalized hysteresis loops in the transverse field geometry for an array of rectangular particles as function of a maximum applied field $H_{\text{max}}$, which has been varied from 212 Oe to 997 Oe. The arrow in Fig. 5-7 is a guide to how the shape of the hysteresis loop is changed at low field as function of $H_{\text{max}}$. For small $H_{\text{max}}$ (< 300 Oe), a continuous decrease of the magnetization near zero field is visible. In contrast, for high $H_{\text{max}}$ (> 300 Oe) the hysteresis loops show a sharp drop near $H=0$, indicating an abrupt nucleation or growth of reversed magnetization. In Fig. 5-8 the normalized magnetization is plotted at applied fields of 50 Oe, 75 Oe, and 100 Oe as function of $H_{\text{max}}$. For $H_{\text{max}}$ smaller than 300 Oe, at all three field values investigated, a drop of up to 25 % of the saturation magnetization has been observed whereas for $H_{\text{max}}$ larger than approximately 300 Oe only a small (10 to 15%) deviation of the magnetization from full saturation is measured. All curves show a transition at around $H_{\text{max}} = 300$ Oe. This
Fig. 5-6 MFM images at H=0 of 0.5 μm wide rectangular and needle-shaped (110) Fe particles after a) and c) longitudinal saturation and after b) and d) transverse magnetic saturation.
Fig. 5-7 Normalized hysteresis loops (MOKE) of a 2 μm wide rectangular particle in a transverse field geometry where the maximum applied magnetic field $H_{\text{max}}$ was varied. The arrow indicates the evolution of the shape of the hysteresis loop with increasing maximum applied field $H_{\text{max}}$. 
transition marks a fundamental change in the magnetization reversal process from a
continuous magnetization reversal process \( (H_{\text{max}} < 300 \text{ Oe}) \) consisting of domain wall
motion, to one \( (H_{\text{max}} > 300 \text{ Oe}) \) driven by a sudden nucleation or growth of reversed
domains in the range of 40 Oe.

MFM images show that, for applied fields smaller than 300 Oe, small domains
remain close to the long edges of the particle and expand during the reversal process.
The MFM images in Fig. 5-9 a) at an applied field of 100 Oe and b) at 50 Oe were
both taken after first applying a transverse field of 900 Oe. These images indicate the
presence of small reversed domains. With decreasing applied field these small
domains become slightly larger. However, at these fields the amount of reversed
magnetization is negligible in comparison with that parallel to the applied field, in
accordance with the data in Fig. 5-8 for \( H_{\text{max}} \) larger than 350 Oe. MFM measurements
were performed on the same particle after applying a maximum field of only 230 Oe.
MFM images at 100 Oe [Fig. 5-9 c)] show that the reversed domains are present on
both long edges of the particle. At 50 Oe [Fig. 5-9 d)] a large reversed domain appears
in conjunction with flux closure domains parallel the long edges of the particle.

These results suggest that pinning of the magnetization at the particle
boundaries depends on the maximum applied transverse field and that this pinning
affects the low-field magnetization reversal. This pinning may be associated with
structural disorder at the particle boundaries caused by the ion-milling process used to
form these structures. Within this scenario, large applied transverse fields more
effectively pin domain walls close to the particle edges leading to the sudden magneti-
Fig. 5-8 Variation of the normalized magnetization at 50 Oe, 75 Oe, and 100 Oe as a function of the maximum applied field $H_{\text{max}}$. 
Fig. 5-9 MFM images of the nucleation process in a 2 μm wide rectangular particle in transverse field geometry. The magnetic domain configuration at a) 100 Oe and b) 50 Oe are shown after applying a maximum field of $H_{\text{max}}=900$ Oe as well as at c) 100 Oe and d) 50 Oe after applying a maximum field of $H_{\text{max}}=230$ Oe.
zation jumps observed at low fields.

5.3.2 Magnetic easy axis parallel to the long particle axis

If the long axis of the particles is positioned parallel to the easy magnetocrystalline axis the particles with rectangular, triangular, and needle-like ends show a single domain state in zero applied field. Here the effective uniaxial anisotropy, $K_{\text{eff}}$, results from both the magnetocrystalline and the shape anisotropy. In the following the z-axis is parallel to the long axis and the x-direction perpendicular to the long axis of the particles. According to a coherent magnetization rotation, the projection of the switching field onto the x- and z-axis is an astroid-like curve (see Chapter 1). The effective uniaxial anisotropy energy density can be described via $E = K_{\text{eff}} \sin^2 \Theta$ with the angle $\Theta$ between the z-axis (particle's long axis) and the magnetization. Minimization of the system's free energy leads to the projection of the switching fields $H_x$ and $H_z$ in the x-z plane described by:

$$ (H_x)^{2/3} + (H_z)^{2/3} = (2K_{\text{eff}}/M)^{2/3} $$  \hspace{1cm} (5-1)

Hysteresis loops have been performed on particle arrays with all three types of ends as function of the angle $\Theta$. The insets of Fig. 5-10 show the hysteresis loops measured on an array of particles with needle-shaped ends at $\Theta=0$ deg, $\Theta=50$ deg, and $\Theta=90$ deg. With increasing $\Theta$ the hysteresis loops evolve gradually from a typical easy axis loop to a hard axis loop. From these kind of hysteresis loops the $H_x$ and $H_z$ values have been determined and plotted in Fig. 5-10 for particles with rectangular (solid rectangles), triangular (solid triangles), and needle-shaped (solid circles) ends. For the
hard axis scan (Θ=90 deg) the H_x values do not show a strong dependence on the exact shape of the particles. This is expected since the rotation of the magnetization will be largely uniform (coherent) across the particle, against magnetocrystalline and shape anisotropy factors. The H_x values for rectangular and needle-like particles have been determined to be H_{x,R}=780 Oe and H_{x,N}=890 Oe, respectively. Magnetocrystalline anisotropy alone is expected to give an anisotropy field of H_a=2(K_u+1/4K_i)/M=540 Oe. The larger values observed are thus associated with element shape anisotropy, which is largest in needle-like particles. Using the measured H_x values of rectangular and needle-shaped particles, astroid curves for both particle types can be plotted. The dashed astroid curve in Fig. 5-10 describes ideal coherent magnetization reversal in the rectangular particles and the dotted one is for needle-like particles. The strong deviations for Θ=0 deg (easy axis loop) from the astroid curves indicate that the reversal is not coherent. In the case of the rectangular and triangular particles reversed domains likely nucleate at the particle ends, as we have observed for particles oriented perpendicular to the [001] axis. The reversal process is driven by this nucleation and subsequent domain wall movement. However, highly metastable magnetization states can be stabilized by blocking the nucleation of magnetic domains at the particle ends. This occurs in needle-like ended particles and results in behavior qualitatively similar to that predicted in a coherent rotation model. Again, this is due to the locally enhanced shape anisotropy for the outermost part of the needles, which suppresses the nucleation of reversed domains. This result is rather remarkable in that large elements, with dimensions much larger than the exchange length (for Fe, δ_x ≈ 20 nm), can show
Fig. 5-10 Magnetization reversal of 2 μm wide rectangular (solid rectangles), triangular (solid triangles), and needle-shaped (solid circles) particles with uniaxial magnetic anisotropy, i.e. the easy magnetocrystalline axis parallel to the long wire axis. $H_z$ and $H_x$ refers to the projection of the in-plane switching field parallel and perpendicular to the long wire axis, respectively. The insets showing the hysteresis loops as function of the magnetic field direction with respect to the long axis of the wire.
characteristics similar to nanometer scale single domain particles. This illustrates clearly the importance of end geometry on the reversal modes of micron scale thin film elements.

5.3.3 Micromagnetic simulations

Micromagnetic calculations have been performed on these Fe particles, particularly on 0.5 μm rectangular particles, in which such calculation are feasible (larger elements require more memory and computational power than presently available). The simulation uses the Landau-Lifschitz-gilber (LLG) Micromagnetics Simulator [5-10]. The equilibrium magnetization is found by solving the dynamic LLG equation,

$$\frac{\partial \vec{M}}{\partial t} = \frac{\gamma}{1 + \alpha^2} \left( \vec{M} \times \vec{H}_{\text{eff}} \right) + \frac{\gamma \alpha}{1 + \alpha^2} \left( \vec{M} \times \vec{M} \times \vec{H}_{\text{eff}} \right)$$  \hspace{1cm} (5-2)

The first term at the right side describes the gyromagnetic motion (precession of the magnetization $\vec{M}$ about the effective field $\vec{H}_{\text{eff}}$), and $\gamma=17.6$ MHz is the electron gyromagnetic ratio. The second term is a damping term, and describes the rotation of the magnetization toward the direction of the effective magnetic field.

The magnitude of $\vec{M}$ remains unchanged by this equation, since the motions generated by both terms in the equation are perpendicular to the magnetization orientation. Only the second term yields an energy change in the system. As we are interested in quasistatic domain patterns, a strong damping constant $\alpha=1$ is used. The effective magnetic field $\vec{H}_{\text{eff}}$ is defined by
\[ \overline{H}_{\text{eff}} = \frac{\partial E}{\partial M} \] (5-3)

where the total energy \( E \) includes Zeeman, magnetostatic, exchange, and magnetocrystalline energy terms.

The complex magnetization domain patterns and the detailed spin structures with the domain boundaries are the results of minimizing the total free energy. That is, the different domain patterns correspond to different local energy minimal. If an external field is applied with sufficient magnitude that the energy minimum disappears, the corresponding magnetization domain pattern will change, following the dynamic equation until a new energy minimum is reached.

The magnetization distribution is approximated by a cubic mesh with a cell size \((10 \text{ nm})^3\). In some cases cells were removed at each of the four corners to mimic shape imperfections found experimentally. Tests performed using finer meshes gave similar results. Experimental value of cubic anisotropy constant \( K_1 = 6.3 \times 10^5 \text{ erg/cm}^3 \) and in-plane uniaxial anisotropy constant \( K_u = 3.1 \times 10^5 \text{ erg/cm}^3 \) are used in the simulations, as well as the values of saturation magnetization \( M_s = 1714 \text{ emu/cm}^3 \), and exchange stiffness \( A = 2.1 \times 10^{-6} \text{ erg/cm} \) of bulk iron. The convergence criterion used was based on the largest allowed angular variation of the magnetization of a single cell during two consecutive iterations. A value of \( 10^{-6} \) was found to be satisfactory.

The hysteresis loops for both longitudinal and transverse field orientations have been calculated. The initial state was chosen to be uniformly magnetized along the positive field direction. The quasi-equilibrium magnetization was then calculated
for a series of fields within a hysteresis loop.

Figure 5-11 shows the simulation result for the transverse case. Above 1 kOe, the magnetization is mainly parallel to the field and lies perpendicular to the long axis of the particles. However, this is not a single domain state. Along the two long sides of particle, the magnetization is tilted towards the edges to reduce the magnetostatic energy. Since these two edge domains are very stable, when the field is decreased to about 750 Oe, a reversed domain is nucleated from the middle of the particles. This process gives rise to a flux-closure structure which contains two vortices. As the field is reduced further through zero field the reversed domain progressively expands towards the short edges of the particles. This process is reversible. When the field reached around -1200 Oe, the two vortices are expelled from the particles which attains near saturation in the negative direction.

For the longitudinal case (Fig. 5-12), end domains appear at the short edges of the particle arranged in the so-called S state, with both end domains pointing in the same direction. Two small reversed domains are also nucleated at the corners. Whilst the center of the particle is still largely parallel to the applied field, both ends of the particle break into a partially flux-closure structure including a vortex. A magnetic vortex configuration at the corners of the particles leads to a minimization of the amount of free magnetic charges and at the cost of exchange and magnetocrystalline energies. These two vortices propagate towards the center of the particle. At H~100 Oe, the nucleation of two other vortices is associated with a strong decrease of the total magnetization. Once this low-field flux-closure structure is nucleated, it evolves
Fig. 5-11 Simulated hysteresis loop for field applied transverse to the particle long axis. In the inset are shown domain structures for several representative values of the applied field. Black arrows give an indication of the magnetization direction.
Fig. 5-12 Simulated hysteresis loop for field applied along to the particle long axis. In the inset are shown domain structures for several representative values of the applied field. Initial nucleation state shows a S state.
in a reversible way when the field is reversed, as four reversed domains extend from both sides toward its center. They collapse for $H \sim -350$ Oe, when nearly negative saturation is achieved.

These calculations are in good qualitative agreement with experiment. The magnetization curves exhibit the same features as the MOKE loops shown in Fig. 5-2. The domain patterns during the reversal are also similar to what observed by MFM. Note that the hysteresis observed experimentally is not perfectly reproduced by the calculation (particularly in the transverse case). This discrepancy is most likely due to a slightly misorientation of the applied field in the MOKE experiments and to some shape irregularities within the array. In the transverse case, the zero-field domain structure (Fig. 5-11) is consistent with the MFM picture shown Fig. 5-6. However, in the longitudinal case, the simulated domain structure of the remanent state does not agree with that found experimentally. The experimental domain pattern of Fig. 5-6 contains two central domains, whereas the calculated structure involves three central domains.

There are several possible reasons why the simulations may not agree with the experimental results. First, there are clearly differences among nominally identical particles in an array such as roughness of the edges, shape of the corners, and locations of the imperfections. All of these influence the amplitude of the nucleation field, as well as details of the magnetization reversal, which in turn influence the domain structure at $H=0$. The MFM image of the magnetic domain configurations of four identical rectangular (110) Fe particles shown in Fig. 5-13 were performed after longi-
tudinal saturation. Three particles contain two stripe domains whereas one exhibits a
domain configuration with only one stripe domain. This MFM image has been
repeated many times each after longitudinal saturation but no change of the
individually domain configuration of a given particle has been observed. This suggests
that the exact domain configuration of an individual particle seems is more strongly
influenced by the specific particle properties than the stochastic nature of the
magnetization reversal.

To clarify the influence of the particle shape, we intentionally introduce some
imperfection into the particles. Figure 5-14 shows MFM images of a) a rectangular
particle, b) a rectangular particle with a cut corner on the lower left-hand side, and c)
with cut corners on the lower left- as well as the upper right-hand side. All MFM
images are performed in a longitudinal oriented field of 300 Oe, which is just below
the nucleation field. For identification, the approximate magnetization direction of the
domains is labeled with arrows. At both sides of all three different particles a magnetic
vortex configuration is visible. The rectangular particle in Fig. 5-14 a) shows a vortex
with right-hand chirality on both ends and therefore both vortex centers are located on
the same long edge of the particle. The magnetizations of two edge domains are
pointing in the opposite direction to form a so-called C state. The MFM images in Fig.
5-14 b) and 5-14 c) demonstrate how sensitive the symmetry of the vortex state is to
slight changes of the particle geometry. By cutting one corner or two corners on
opposite sites, the center of the vortices appears on the side with the remaining
rectangular corner geometry. In both cases the vortices in each particle are of opposite

136
Fig. 5-13 MFM image at $H=0$ of an array of 0.5 $\mu$m wide rectangular $(110)$ Fe particles after longitudinal magnetic saturation.
Fig. 5-14 MFM images of magnetic vortices in a) a rectangular particle, b) a rectangular particle with one cut corner, and c) a particle with two cut corners at an applied field of $H=300$ Oe. Domain structure in a) shows a C state, while domain structure is c) is an S state.
chirality. This leads to an S state of edge domains [Fig. 5-14 c]).

Micromagnetic simulation also shows that modifications of the element's shape near one of its corners allows the stabilization of the C state in a 0.5 μm wide rectangular particle and nucleation of a two-central-domain zero-field pattern (Fig. 5-15), which was observed in experiment.

The discrepancy can also be understood in terms of the metastability of the nucleation process of the reversed domains. As shown in Fig. 5-12, when a large field is applied along the particle axis, the S state is stabilized. Given this S state nucleation pattern, we obtain the zero-field pattern shown in Fig. 5-12. However, under certain conditions, it is also possible to stabilize the C state. This C state was obtained in simulation by applying a moderate longitudinal field H~600 Oe, starting from a random initial state (Fig. 5-16). For the C state pattern, the remanent domain structure exhibits only two central domains, in good agreement with experiments. This domain structure has the lowest energy of $E=1.53\times10^{-8}$ erg, compared to $E=1.58\times10^{-8}$ erg for the three-central-domain pattern and $E=1.74\times10^{-8}$ erg for the one-central-domain pattern obtained after transverse saturation, but also shown in Fig. 5-13 after longitudinal saturation.

To summarize this chapter, micron scale (110) Fe elements have been used as model materials to investigate the effect of shape and competing anisotropies on magnetization reversal and micromagnetic configurations. Domains are nucleated at corners in rectangular particles, while in the central part of needle-shaped particles. Trapped domains near edges effect the magnetization reversal in the transverse case.
Individual particle properties (geometry, defects, etc.) play an important role in the nucleation of a flux-closure domain structure in zero field.
Fig. 5-15 Defects at the particle corner allows the stabilization of the C state (left image, $H=600$ Oe), and leads to two central stripe domains at $H=0$ (right image).

Fig. 5-16 Starting from a random initial state, C state was obtained in a perfect rectangular particle by applying a moderate longitudinal field $H \sim 600$ (left image), and leads to two central stripe domains at $H=0$ (right image).
Chapter 6 Exchange Biasing in Polycrystalline Thin Film Microstructures

The exchange bias effect results from exchange coupling between adjacent ferromagnetic (FM) and antiferromagnetic (AFM) layers after these layers are either deposited in a magnetic field or cooled down in a magnetic field from above the Neel temperature. The hysteresis loop of these layers is not symmetric about zero applied magnetic field, but exhibits an offset from zero field, which is defined as exchange biasing field, and it is often described as an unidirectional anisotropy. The second phenomenon usually observed is an enhancement of the coercivity of the ferromagnetic layer. Exchange biasing is used to control the magnetization in devices, such as spin valves which sense changing magnetic fields through GMR effect. The exchange biasing at FM/AFM interface hardens one FM layer while the other FM layer is free to rotate in an applied magnetic field. Although the exchange biasing is of critical importance to magnetic memories and sensors, mechanisms of exchange biasing are poorly understood.

In this chapter I present the experimental studies of the effect of lateral element sizes, shape and FM layer thickness on the EB field and coercivity in polycrystalline thin film magnetic elements comprised of CoFe/IrMn. Such materials have been used as the fixed layer in high MR magnetic tunnel junctions, as IrMn has
both a large pinning field and high blocking temperature.

6.1 Models of exchange biasing

The simplest model [Section 1.1.4] predicts an exchange field of order \( J_i/a^2 M_F t_F \), in which \( J_i \) is interfacial exchange integral constant, \( a \) is lattice size, and \( M_F, t_F \) are the magnetization and thickness of the ferromagnetic layer. But this predicted result is orders of magnitude too strong compared to the loop shifts that have been measured. Mauri et al. [6-1] considered the consequences of the formation of a DW parallel to the interface in the antiferromagnetic layer (Fig. 6-1). The exchange energy is now spread out over a domain wall width \( \sim \pi (A_A/K_A)^{1/2} \), which dramatically lowers the energy required to reverse the magnetization. This model predicts an exchange field

\[
H_E = 2\sqrt{A_A K_A} / M_F t_F
\]  

(6-1)

where \( A_A \) and \( K_A \) are the uniaxial anisotropy energy and exchange stiffness of the AF layer, \( M_F \) and \( t_F \) are the magnetization and thickness of the FM layer. With typical parameters, \( t_F H_E \) is around \( 10^5 \) Oe Å, which is in better agreement with experiment.

While previous models consider a perfect "uncompensated" interface where the moments of the first layer of the antiferromagnet in contact with ferromagnetic layer are ferromagnetically aligned, Koon [6-2], as well as Schultess and Butler [6-3] has studied ideal interface that are completely compensated, i.e. an interface with no net moment. The model showed spin-flop coupling, that is the FM spins align perpendicular to the AFM spins at the interface (Fig. 6-2). This is the lowest energy
state of an AFM in a magnetic field. They find that this type of coupling leads to a
uniaxial anisotropy which can increase the coercivity. However, they find that an
ideal compensated interface does not lead to EB. Introducing defects, that is a degree
of uncompensation, is found to be necessary to produce EB.

Malozemoff has taken a different approach. He considered the likely
randomness in exchange interactions at the interface, arising from surface roughness
or chemical inhomogeneity on an atomic scale. This random exchange across the
interface will break the AF to domainlike regions, as pictured in Fig. 6-3, to
minimize the system energy. The AF domain size $L$ is around domain wall width
$\pi(A_{x}/K_{A})^{1/2}$. Once these domains are fixed, flipping the ferromagnetic orientation
causes an energy change, which gives the exchange bias field

$$ H_{E} \propto \sqrt{A_{x}/K_{A}}/M_{F} t_{F} $$

(6-2)

This randomness may also break the FM layer into small domains, and DWs can be
trapped into local energy minimum, increasing the coercivity [6-5].

These models commonly suggest the DW formation or the domain structure
in the AFM layer affects the magnetic properties of the exchange-coupled layers.
The nature of EB is expected to change significantly as the AF/FM element sizes
approach these domain sizes. Next, I will describe the experiments on
lithographically defined IrMn/CoFe particles.

6.2 Experiment

A series of polycrystalline thin films were fabricated by magnetron sputtering
Fig. 6-1 Mauri’s model for the interface of a thin ferromagnetic film on a thick antiferromagnetic substrate. The uniaxial anisotropy of the antiferromagnet is along the Z-axis. The figure depicts a situation in which an external magnetic field is applied opposite to Z and in which the exchange coupling across the interface is positive.
Fig. 6-2 Spin configurations near the interface for a FM/AFM film.

Fig. 6-3 Schematic perspective view of antiferromagnetic domains.
on oxidized silicon substrates. The structures consist of Si/SiO<sub>2</sub>/Ti (5nm)/Pd (15 nm)/IrMn (12 nm)/CoFe (t<sub>c</sub>)/Al<sub>2</sub>O<sub>3</sub> (12 nm), with CoFe thickness, t<sub>c</sub>, equal to 2, 2.5, 3, and 5 nm. Single CoFe film without IrMn layers were also produced for comparison. These films have been patterned into particle arrays similar to our Fe particles (Fig. 5-1), using electron beam lithography and ion milling. The long axis of the particle is oriented parallel to the exchange biasing direction. Magnetization reversal processes are probed by combining longitudinal magneto-optical Kerr hysteresis loop (MOKE) measurements of arrays with magnetic force microscope (MFM) imaging of individual array elements. In MOKE measurements the applied field has been cycled many times to obtain a single (averaged) hysteresis loop and to eliminate training effects. Both average H<sub>e</sub> and H<sub>c</sub> values are deduced from such measurements. MFM imaging is performed using vertically magnetized NiFe coated tips.

6.3 Results

Figure 6-4 shows typical hysteresis loops of both exchange-biased and unbiased reference samples. Compared to unbiased samples, the exchange-biased samples show loop shifts away from H=0, and have much larger coercivity. For example, the hysteresis loop of IrMn/CoFe (5nm) film is shifted 213 Oe and the coercivity is 69 Oe. While for unbiased CoFe (5nm) film, the loop is centered at H=0 and coercivity is only 20 Oe.

The coercivities of both unbiased (Fig. 6-5) and exchange-biased (Fig. 6-6)
CoFe particles are plotted as a function of FM layer thickness. For unbiased samples, $H_c$ increases after patterning and is largest in needle-shaped elements. For 5 nm thick 0.5 μm wide rectangular particles, the coercivity is 78 Oe, while the coercivity of needle-shaped particles is 116 Oe. For a particular shape, the coercivity increases with increasing FM layer thickness. For example, the coercivity of 0.5 μm needle-shaped particles varies from 84 to 116 Oe as the CoFe thickness is varied from 2 nm to 5 nm.

In contrast, Fig. 6-6 shows the thickness and size dependence of the coercivity of EB films and patterned elements. Both the coercivity is enhanced and the dependence on element shape has completely changed qualitatively. At 2 nm CoFe thickness, the EB film has a coercivity of 175 Oe, compared to 32 Oe for an unbiased CoFe film. In contrast to the unbiased films, $H_c$ decreases from 179 Oe to 69 Oe with increasing FM film thickness, from 2 nm to 5 nm. The solid curve shows this trend. Also, $H_c$ is largely unchanged in small patterned elements. While the trends are the same, there is some scatter in the data on patterned elements.

Figure 6-7 shows the EB field and coercive field plotted versus the inverse of the FM layer thickness. By fitting the film data, we have found that both $H_E$ and $H_c$ are inversely proportional to the FM layer thickness $(1/t_f)$. This indicates that the exchange biasing field and coercivity are both dominated by interactions at the FM/AF interface. Note also, $H_E$ is reduced slightly (~10%) in patterned elements.

Brillouin light scattering (BLS) has been used to investigate the possible additional anisotropy contributions induced by the exchange coupling mechanism.
Fig. 6- 4 Typical hysteresis loops observed in CoFe (upper panel) and IrMn/CoFe (lower panel) films and particles (2μm wide).
Fig. 6-5: The coercive field of reference (unbiased) CoFe films and particles.
Fig. 6-6 The coercive field of exchange biased IrMn/CoFe films and particles
Fig. 6-7 The coercive fields (solid symbols) and exchange bias fields (open symbols) of IrMn/CoFe films (circles) and particles versus $1/t_f$. Both $H_E$ and $H_c$ of films scale as $1/t_f$. 

152
The BLS experiments were performed with the external field applied in the film plane and in magnetic saturation to ensure a single-domain state of the FM layer. A laser light was focused onto the sample. The light inelastically scattered from thermally excited spin waves was frequency analyzed. The spin wave frequency is varied with the orientation of the applied saturation field. Usually the spin wave maximum indicates the easy axis of magnetization, while the minimum indicates the hard axis. Figure 6-8 shows the spin wave frequencies as a function of the angle of the in-plane applied field in IrMn/CoFe (5nm) and unbiased CoFe (5nm) films. In CoFe film, the spin wave frequency is almost a constant with the angle $\phi$, which indicate there is no anisotropy at all, consistent with a polycrystalline film. In contrast, the IrMn/CoFe film shows a maximum and a minimum. From this kind symmetry, it can be concluded that there is no additional in-plane uniaxial anisotropy in our exchange biased films. To obtain the anisotropy constants, the following expression for the free energy density was used:

$$E = -K_p \cos^2 \theta + K_u \cos(\phi - \phi_{un}) \sin \theta$$  \hspace{1cm} (6-3)

with $K_u$ the in-plane unidirectional anisotropy constant, $K_p$ the perpendicular anisotropy, $\phi$ the in-plane angle of the direction of magnetization, $\phi_{un}$ describes the reference direction of the unidirectional anisotropy, and $\theta$ is the out-of-plane polar angle. By fitting the experimental data it is found that $K_u = 6.4 \times 10^4$ J/m$^3$. The exchange bias field is calculated to be $2K_u/M_F = 330$ Oe, which is larger than what we found in hysteresis loop. Usually reversible measurements like BLS give higher exchange bias fields than irreversible measurements such as hysteresis measurement.
Fig. 6-8 Spin wave frequency as a function of in-plane angle of the applied field in IrMn/CoFe (5 nm) and CoFe (5 nm) films.
The hysteresis technique is dominated by the irreversible process, such as domain nucleation and propagation, while in the BLS technique the FM is always nearly uniformly magnetized.

MFM imaging has been performed in order to correlate magnetic hysteresis measurements with the magnetization reversal processes and magnetic domain patterns of individual particles. Figure 6-9 shows a magnetization reversal process observed in a rectangular IrMn/CoFe (5-nm) particle. The labeled dots on the sketched loop indicate the conditions at which the corresponding MFM image was taken. Starting from an almost saturated state (H=148 oe), the domains nucleate everywhere inside the particle at H=123 Oe, not only at the corner which is usually observed in unbiased particles [section 5.3.1]. With decreasing field, this leads to fine scale domain subdivision of the FM layer at the coercive field (Fig. 6 -9c). For \( t_r = 5 \text{ nm} \), the particle shows multi-domains with size about 700 nm at \( H_c = 99 \text{ Oe} \). The shape of domains is irregular. Such domain configurations are completely different from that observed in unbiased elements, which typically show a small number and more regularly shaped domains during magnetization reversal. Strong domain wall pinning has also been observed. For example, the most right domain (indicated by the arrow) remains at the same position through all the field range (-99 Oe to 37 Oe). The MFM images show the reversal process took several jumps (from c to e and from e to g). Between c and e, e and g, the domain configuration barely changes with the field.

Similar behavior has been observed in particles with thinner FM layer. Figure
6-10 shows an example in an IrMn/CoFe particle with $t_F = 2$ nm. However, the domain walls become more convoluted, and domain sizes get smaller ($\sim$ 500 nm) at $H_C = 175$ Oe. Depending on whether the applied field is increasing or decreasing, the domain structure is different. Figure 6-11 shows the other branch of the reversal process in the same particle showed in Fig. 6-10. Although the nucleation and domain sizes are similar, domain walls appear at different positions. This is contrast to very asymmetric nucleation observed in epitaxial NiO/NiFe bilayers [6-6]. In that case, when the magnetic field is aligned against the unidirectional anisotropy axis, the nucleation of domains occurs at the film edges. However, when the field is aligned along this axis, the domain nucleation takes place at the sites associated with lattice defects.

It has also been observed that the magnetization reversal processes in EB films are quite similar to those in particles. Figure 6-12 shows an example.

6.4 Discussion

These results show clearly a distinction between patterned exchange biased and unbiased elements. For unbiased samples, patterning generally increases the element coercivity. Whereas in exchange biased samples, patterning (to micron scales) neither significantly alters the exchange bias field, nor the coercivity. In the unbiased elements, with increasing film thickness, decreasing particle size, and increasing element aspect-ratio the demagnetizing fields arising from magnetic charges at particle edges become more important. These introduce a shape
Fig. 6-9 Field dependent MFM images of a rectangular IrMn/CoFe (5 nm) particles.

The applied field is parallel to the unidirectional anisotropy axis.
Fig. 6-10 Field dependent MFM images of a rectangular IrMn/CoFe (2 nm) particles.

The applied field is parallel to the unidirectional anisotropy axis.
Fig. 6-11 Field dependent MFM images of a rectangular IrMn/CoFe (2 nm) particles.

The applied field is parallel to the unidirectional anisotropy axis.
Fig. 6-12  A MFM image of a IrMn/CoFe (2 nm) film at $H = -460$ Oe. It is similar to what observed in Fig. 6-11 e).
anisotropy with easy axis parallel to the long axis of the particle and leads to an increase in the coercivity. In exchange biased elements, evidently this shape anisotropy does not play as important a role.

The increase in coercivity in both the exchanged biased films and elements is dominated by interactions between the AF and FM film. Models which find spin-flop coupling between the FM and AF layers also show that this coupling induces a uniaxial anisotropy, which would increase the coercivity [6-2]. But the BLS investigation of CoFe/IrMn films shows the in-plane unidirectional anisotropy, with no apparent induced in-plane uniaxial component to the anisotropy. Based on these results and our MFM studies we infer that the increased coercivity in these IrMn/CoFe films and elements is associated mainly with changes in the magnetization reversal process.

Zhang [6-5] has showed that the random interaction at the interface breaks the FM into domains during the reversal process. For a small FM thickness (e.g. <8 nm), the domain size is around the film grain size, and each domain receives different random fields from the AF layer. Due to fluctuation of the random fields from each domain, the reversal will take place at different external fields for different ferromagnetic domains. Therefore the breakup of the FM layer into small domains is essential of the increased coercivity. Since the domain size is limited by the grain size, further reduction of the thickness does not increase the coercive energy from the random field. Therefore, it is reasonable that the coercivity scales with the F layer thickness as $H_C \propto t_F^{-1}$, which is what we observed in our experiments.
We have observed that in IrMn/CoFe samples small domains are nucleated "randomly" in the interior of elements and that domain walls are strongly pinned. To move these domain walls the energy barriers determined by local fluctuation of the random fields in the film need to be overcome. This may introduce the jumps in DW motion observed by MFM. Assuming the random field is spatially uncorrelated, the domain patterns should be essentially uncorrelated between two branched of the hysteresis. Similar results have been observed in exchange biased NiFe films by transmission electron microscope (TEM) [6-7,8] and the magneto-optic indicator technique [6-6]. Hysteresis loop measurements average over an array of particles, which smooth out the details of the magnetization reversal in a single particle.

In conclusion, the driving force for the domain nucleation and the subsequent impediments to domain wall motion is those associated with the exchange interactions which vary in the plane of the AF/FM interface. The scale of these fluctuations are much smaller than the size of our patterned elements, since similar magnetization reversal and magnetic characteristics (H_c and H_e) are observed in our films and patterned structures. Element size effect may become prominent when it approaches characteristic magnetic lengths (e.g. the exchange length ~ 50 nm) and grain sizes in these materials.
Chapter 7 Summary

Through my thesis work, I have studied the effect of size, shape and boundaries on the magnetic and electronic transport properties of lithographically patterned thin film structures. These include domain wall scattering of conduction electrons in bcc Fe, hcp Co, and L1₀ FePt, as well as magnetization reversal and domain configurations in epitaxial Fe and polycrystalline CoFe/IrMn elements. These studies have involved a variety of techniques, such as advanced thin film deposition methods, nanofabrication, magnetotransport measurement, magnetic characterization via MOKE and MFM, and micromagnetic simulations.

For domain wall scattering, I found that micromagnetic structure and resistivity anisotropy are the predominant sources of low field magnetoresistance in both Fe and Co. Domain walls have a smaller effect on MR. In Fe the presence of domain walls reduce the resistivity, which may be intrinsic or due to “domain” effects, such as the influence of spatially varying internal fields on electron trajectories which act to reduce scattering at film surfaces. In hcp Co, the difference between CPW and CIW resistivities provides a means of estimating DW-MR within a GMR-like mechanism of domain wall scattering. Very high anisotropy in L1₀ FePt films leads to well defined domain configurations with the magnetization perpendicular to the current, hence, small AMR contribution. In FePt film with highest anisotropy, evidence for an intrinsic DW scattering contribution has been
found, and domain walls enhance the resistivity. However, an effect associated with Hall effect might explain observed MR in lower anisotropy films.

These experiments, their analysis and interpretation suggest interesting directions for future research. Since DW scattering effects are predicted to depend strongly on DW width, it would be interesting to extend such studies to films with even higher magnetic anisotropies. Another way to explore the DW scattering is study nanometer scale structures, such as magnetic nanocontact. It has been reported that domain wall scattering in such a system may produce 300% MR effect [7-1,2]. There are some models, which predict that DW MR depends on asymmetry of spin dependent scattering. To test such models, different impurities could be intentionally doped into samples to change the spin dependent scattering. The DW resistance as a function of type and concentration could be measured.

Micron scale (110) Fe and polycrystalline CoFe/IrMn elements have been fabricated. Direct imaging in applied fields with MFM combined with hysteresis loop measurements of arrays lead to a detailed understanding of the magnetization reversal. In Fe particles, end shape and individual particle properties (geometry, defects, etc.) are important factors in the magnetization reversal process. Micromagnetic simulations well describe the magnetization reversal mode and domain configurations observed in Fe elements. To get better performance in real magnetic devices, it is thus quite important to control the shape of the elements. However, it is much more challenging to control the defects at edges to produce uniform switching behavior.
Pattern CoFe/IrMn films to micron sizes does not significantly alter exchange biasing phenomena. We found that coercivity $H_c$ is inversely proportional to the ferromagnetic layer thickness, and there is no additional in-plane uniaxial anisotropy. Magnetization reversal occurs via formation of small domains, with strong pinning of domain walls. Our results suggest that randomness of exchange interaction at interface is the essential component to understanding the phenomena. To observe clear size effects, it will be necessary to fabricate elements with lateral sizes down to deep submicron level, which approaches the domain sizes in the antiferromagnetic or ferromagnetic layers.
# Appendix I  Lithography Process Notes

<table>
<thead>
<tr>
<th></th>
<th>10x Stepper</th>
<th>Contact Aligner</th>
<th>E-beam</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Resist Type</strong></td>
<td>OCG OiR 897-7i</td>
<td>Shipley 1813</td>
<td>SAL-603 2:1 with thinner p</td>
</tr>
<tr>
<td><strong>Substrate</strong></td>
<td>Spray acetone and IPA while spinning; P-20 liquid prime</td>
<td>Spray acetone and IPA while spinning; P-20 liquid prime</td>
<td>Spray acetone and IPA while spinning; YES Oven HMDS vapor prime</td>
</tr>
<tr>
<td><strong>Spin Resist</strong></td>
<td>4000 rpm, 40 seconds, ~1 μm</td>
<td>4000 rpm, 30 seconds, ~1.3 μm</td>
<td>3000 rpm, 60 seconds, ~470 nm</td>
</tr>
<tr>
<td><strong>Bake Resist</strong></td>
<td>90 °C, 1 minute, hot plate</td>
<td>90 °C, 1 minute, hot plate</td>
<td>105 °C, 1 minute, hot plate</td>
</tr>
<tr>
<td><strong>Expose</strong></td>
<td>depends on structures</td>
<td>depends on structures</td>
<td>depends on structures</td>
</tr>
<tr>
<td><strong>Post-exposure bake</strong></td>
<td>115 °C, 90 sec, hot plate</td>
<td>90 °C, 1 minute, hot plate</td>
<td>110 °C, 1 minute, hot plate</td>
</tr>
<tr>
<td><strong>Develop Resist</strong></td>
<td>OCG OPD 262 (no dilution), 1 minute</td>
<td>MF-312 (1:1 with DI water), 1 minute</td>
<td>CD-30, 4 minutes</td>
</tr>
<tr>
<td><strong>Ion Mill</strong></td>
<td>Ar⁺, 500 V, 100 mA, time depends on film thickness</td>
<td>Ar⁺, 500 V, 100 mA, time depends on film thickness</td>
<td>Ar⁺, 500 V, 100 mA, time depends on film thickness</td>
</tr>
<tr>
<td><strong>Strip Resist</strong></td>
<td>1165 remover with ultrasonic agitation</td>
<td>1165 remover with ultrasonic agitation</td>
<td>1165 remover with ultrasonic agitation</td>
</tr>
</tbody>
</table>
Appendix II Image reversal

5-1 Liquid prime with P-20 (20% HMDS) primer. Apply primer over entire wafer, allow it to remain for 10 seconds, then spin dry (3000-5000 rpm, 30 sec).

5-2 Coat photoresist at desired speed for 20-30 seconds.

5-3 Solvent removal bake at 90 °C for 1-2 minutes on the hot plate.

5-4 Expose. Time will vary depending on resist thickness, bake time, substrate reflectivity, etc.

5-5 Run YES Oven Ammonia diffusion process.

5-6 Flood expose for 30 seconds using the HTG contact aligner.

5-7 Develop for 1 minute in MF 321 (no dilution).
References


1-3. Alex Hubert and Rudolf Schafer, Magnetic Domains (Springer-Verlag, 1998).


5-7 C. Kittel, Phys. Rev. 70, 965 (1946).


