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Magneto-optical Kerr Effect Analysis of Magnetic Anisotropy in Soft Ferromagnets

by

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A dissertation submitted in partial fulfillment of the requirements for the degree of Master of Science in Physics
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Abstract

The continued progress of modern information technology relies on understanding the influence of magnetic anisotropy on magnetic thin films. In this work, two sources of magnetic anisotropy are examined in two different soft ferromagnets: a uniaxial anisotropy induced during the fabrication of Ni$_{80}$Fe$_{20}$ and exchange anisotropy, or exchange bias, which occurs at the interface of Ni$_{77}$Fe$_{14}$Cu$_{5}$Mo$_{4}$/Fe$_{50}$Mn$_{50}$ bilayer. A home-built Magneto-optical Kerr effect magnetometer is used to measure the magnetic response of the soft ferromagnetic films and details of its construction are also discussed. A simple model of uniaxial anisotropy is described, then applied, to the uniaxial NiFe film and deviations from the model are critically analyzed. The exchange bias and coercive fields of NiFeCuMo/FeMn are reported for the first time and studied as a function of buffer layer material. The influence of the different buffer layer materials on the magnetization response of the bilayer is explained from a structural standpoint.
1 Introduction

The magneto-optical Kerr effect (MOKE) has been applied to the study of technologically relevant magnetic materials for the past thirty years [1]. The elegance, simplicity, and non-destructive nature of MOKE make it a standard technique for measuring the magnetization response of a magnetic material. This response, usually a hysteresis loop, provides a characteristic fingerprint of the material under study. In this work, MOKE was used to characterize soft ferromagnetic materials under the influence of different types of magnetic anisotropy. Understanding the influence of magnetic anisotropy in magnetic thin films is important to the continued progress of modern information technology as soft anisotropic thin films are a vital component to devices such as the hard disk drive.

The phenomenon of MOKE was discovered by Reverend John Kerr in 1877. Earlier in that century crucial discoveries were made in order to better understand the interaction of light and other known physical forces. For example, in 1845 Michael Faraday found that the polarization of light would rotate as it passed through a substance magnetized by a magnetic field. The work of Faraday would pave the way for Kerr to examine the effect on light upon reflection from a magnetized surface. Kerr indeed found that a beam of light, initially plane polarized, would exhibit a rotation of the plane of polarization upon reflection from a polished mirror of magnetized iron. He was awarded the Royal Medal in 1898 for his research, upon which the presenter of the medal remarked in wonder how it could be that Kerr learned so much from the “comparatively simple and ineffectual apparatus...” Kerr then replied,
quite fittingly, “Simple it may be, but not ineffectual; rude, but not crude.”

The statement made by Kerr certainly echoes a theme in this thesis. The majority of the magnetic measurements presented in this work were made by a simple and inexpensive, yet effective, home-built MOKE magnetometer. For example, the support for the optical components of the magnetometer was constructed from an old and heavy wooden desk top cut into an “L”-shape. The sample holders were made from store-bought LEGOs and it is even possible to use an off-the-shelf laser pointer as the light source. More details about the construction of an inexpensive MOKE magnetometer are given in Section 3.3.2.

The remainder of this thesis is organized as follows: In Section 2, the reader is introduced to important concepts in thin film magnetism. Section 3 explains the experimental methods used to fabricate and characterize the thin film samples, including construction details of the aforementioned MOKE magnetometer. Finally, Section 4 presents the results and discussion of two studies on anisotropic magnetic thin films.
2 Magnetism in Thin Films

2.1 Magnetic Energy Contributions

2.1.1 Exchange Energy

The exchange interaction in magnetism is a quantum mechanical phenomenon that causes cooperative magnetic ordering. When an individual magnetic moment attempts to align with its nearest neighbors, the energy due to the exchange interaction is expressed as:

$$E_{\text{exch}} = -2J \sum_{i<j} S_i \cdot S_j$$  \hspace{1cm} (1)

where $J$, the exchange constant, represents the strength of the interaction between the moments and $S_{i,j}$ is unit vector of the spin; proportional to the magnetic moment. The summation is carried over all the nearest neighbors of moment $i$. If $J$ is positive, the preferred alignment is parallel, and the material is said to be ferromagnetic. If $J$ is negative, the moments prefer to align antiparallel, and the material is antiferromagnetic.

2.1.2 Magnetostatic Energy

The magnetostatic energy depends on the magnetization, $M$, that arises from the alignment of magnetic moments in a material. It can be understood as the energy generated by the magnetic body, or self-energy, and is given by:

$$E_M = -\frac{1}{2} \int_V \mathbf{M} \cdot \mathbf{H}_d dV$$  \hspace{1cm} (2)
where $H_d$ is the demagnetizing field, which is the field produced by the magnetization of the magnet. It is evident from Eq. 2 that the magnetostatic energy has long range character, since the expression depends on the shape of the magnetic body. In fact, it is the minimization of magnetostatic energy which leads to a type of anisotropy called shape anisotropy. Magnetostatic effects also have great influence over the domain wall structure in thin films. [2]

2.1.3 Magnetic Anisotropy Energy

Magnetic anisotropy, which is the preference for the magnetization to lie in a particular direction in space, can originate from many sources. The sources of anisotropy relevant to this work are due to sample shape, crystal structure (magnetocrystalline), or are magnetic field induced. The symmetry of the anisotropy is considered when forming the anisotropy energy term. Each type of anisotropy mentioned can have symmetry about a single axis. This axis could be a crystallographic axes, such as the c-axis in Co, or it could be the axis parallel to an applied magnetic field during deposition of a sample. Anisotropy energy with uniaxial symmetry has the form

$$E_{ani} = K_1 \sin^2 \theta$$

(3)

where $K_1$ is the uniaxial anisotropy constant and $\theta$ is the angle between the magnetization and the anisotropy axis or easy axis. Another relevant source of anisotropy to this work is exchange anisotropy. It is the anisotropy induced by the coupling of two ordered magnetic materials through a physical interface. In the case of ferromagnetic/antiferromagnetic coupling, also known as exchange bias, the symmetry of the anisotropy is unidirectional. It has anisotropy energy of the form

$$E_{ani} = K_u \cos \theta$$

(4)
where $K_u$ is the unidirectional anisotropy constant and $\theta$ is the angle from the anisotropy axis. The unidirectional nature of exchange bias will be elaborated in the following section.

### 2.1.4 Zeeman Energy

The Zeeman energy is a potential energy of the interaction between the magnetization, $M$, and an external magnetic field. The form of this energy term is simply

$$E_Z = -M \cdot H = -MH \cos \theta. \quad (5)$$

The Zeeman energy is minimized when $H$ is parallel to the external field.

### 2.2 Origins of Magnetic Anisotropy

A material property is said to be anisotropic if it shows a preference to a particular direction in space. In the case of magnetic materials, the preference for the magnetization to lie in a particular direction is known as magnetic anisotropy. The intrinsic source of magnetic anisotropy can be found in the crystal structure of the magnetic material and is called magnetocrystalline anisotropy. The shape or stress of a magnetic material can also be a source of magnetic anisotropy. Additionally, magnetic anisotropy can be induced by placing two ordered magnetic materials in contact, for example, a ferromagnet and an antiferromagnet. This is called exchange anisotropy or exchange bias. The origins of intrinsic magnetocrystalline anisotropy and exchange bias will be presented in greater detail.

#### 2.2.1 Magnetocrystalline Anisotropy

The principle origin of magnetic anisotropy is known as magnetocrystalline anisotropy and originates from the crystal field’s influence on the spin-orbit interaction [3]. Here,
the crystal field is defined as the influence of neighboring spins on the point of interest on the crystal lattice. When an external applied magnetic field tries to reorient the spin of an electron, the orbit of that electron also reorients due to spin-orbit coupling. Considering that the orbit is also strongly coupled to the crystal lattice, there is a resistance to the reorientation. It is this competition between the spin-orbit coupling and crystal field interaction that gives rise to magnetocrystalline anisotropy. Materials with a crystal field that dominates over the spin-orbit interaction have low magnetocrystalline anisotropy. Examples of these are the transition metals (Fe, Ni, Co) and alloys, such as NiFe or Permalloy (Py), where the magnetic moments originate mainly from the spin (and not the orbit) of the 3d electrons. On the other hand, it is the orbital motion of the 4f electrons which contribute to the magnetic moment of the rare earth magnets. These elements have a large magnetocrystalline anisotropy due the small interaction of the 4f orbitals with the crystal field.

It is important to note that a polycrystalline sample will yield, at most, a very weakly defined magnetocrystalline anisotropy due to the random distribution of crystallites. However, it is possible to induce uniaxial magnetic anisotropy in a magnetically soft polycrystalline alloy by applying a magnetic field in the film plane during deposition or by magnetic annealing. The anisotropy axis is likely to exist as a distribution within the material due to the intrinsic distribution of crystallites within the polycrystalline material.

2.2.2 Exchange Bias

Meiklejohn and Bean [6] were the first to observe the manifestation of a coupling of spins on either side of a physical interface. Specifically, their specimen was fine Co ferromagnetic particles that had oxidized at the surface to produce a thin layer of CoO. After heating their specimen and cooling it in the presence of a magnetic field, they measured the hysteresis loop of the particles and found the loop to be shifted
Figure 2.1: *from [5]*. A cartoon depicting the spin configurations of a FM/AF bilayer at different stages of an exchange-biased hysteresis loop.
Figure 2.2: from [7]. A schematic depicting the realistic complexities of a polycrystalline FM/AF interface. The marked x’s indicate uncompensated spins at the interface which can effect the exchange bias field.

along the field axis. Conversely, when no field cooling procedure was performed, the loop was symmetric about the field axis. They traced the evidence of the shifted loop to the intimate contact of the Co particles with the thin layer of CoO, an antiferromagnet (AF). They concluded that they discovered a new magnetic anisotropy, which can be induced when two different magnetic materials are in intimate contact with one another. This discovery, which describes a case of ferromagnetic (FM)-antiferromagnetic (AF) exchange coupling, is known today as exchange bias (EB).

The phenomenological understanding of the mechanism of exchange bias is based on the spin configuration of a FM and AF in close contact at a microscopic level. Let the FM and AF be in a single domain state and the the Néel temperature ($T_N$) of the AF be less than the Curie temperature ($T_C$) of the FM. Figure 2.1 (i) shows that when the temperature of the sample is above $T_N$ but less than $T_C$, the AF is in a paramagnetic state while the FM spins align with the applied magnetic field. As the temperature is lowered through $T_N$, the AF spins at the interface will couple to the FM spins either ferromagnetically (parallel) or antiferromagnetically (antiparallel). In Fig. 2.1 (ii) they are shown to be coupled ferromagnetically. To
complete the antiferromagnetic order; the remainder of the spins of the AF couple to those at the interface antiferromagnetically. (i) and (ii) illustrate the field cooling procedure that is widely used to induce exchange bias in FM/AF systems.

The remainder of Fig. 2.1 examines the spin configurations of the FM/AF system during a hysteresis loop measurement. If the loop begins at the end of the field cool procedure, the FM spins are saturated to a positive magnetic field. When reversing the field, the FM spins try to follow the field but are impeded by the exchange coupling across the interface. This amounts to more Zeeman energy required to rotate the FM spins, resulting in a coercive field (left side) that is larger in magnitude. When applying again a positive magnetic field to bring the system back from negative saturation, the FM spins require less Zeeman energy to return, and the (right) coercive field is reduced in magnitude. The overall result is a shift of the field-cooled hysteresis loop opposite to the field cool direction. This is called more specifically negative exchange bias, as opposed to positive exchange bias, which is when the loop shift is along the field cool direction.

Field cooling, however, is not the only method used to induce exchange bias. Instead, EB was induced in the samples in this work during the sputter deposition process. Underneath the sample plate several permanent magnets are placed such that an approximately 1000e field is produced in the plane of the substrate with negligible field out of the plane. During deposition of the ferromagnet, the spins align to the field, and the FM is deposited with an induced magnetic anisotropy. Next, the antiferromagnet is deposited and each monolayer of deposition aligns to the induced magnetic anisotropy of the FM in the same way described in the field-cool procedure previously mentioned.

The aforementioned phenomenological picture of exchange bias accurately describes the presence of the loop shift and other qualitative features. However, it provides little quantitative analysis and generally overestimates the value of the ex-
change bias field by several orders of magnitude. The reason for this is the model’s
ingability to describe the interfacial complexities of a polycrystalline (or multi-domain)FM/AF interface. These complexities include the role of interfacial roughness, spin
configuration at the interface, and FM or AF grain structure. These properties are
often times interrelated, making a unified exchange bias theory very complicated.

Extensive research is being conducted to better understand the exchange bias
effect. Relevant theoretical and experimental results concerning these efforts are
summarized in the reviews by Noguès and Schuller [5], and Berkowitz [7].
3 Experimental Methods

3.1 Sample Fabrication

Physical vapor deposition (PVD) is a class of vacuum deposition methods used to deposit thin films by the condensation of a vaporized form material onto surfaces of substrates. One of the most popular methods of PVD, magnetron sputtering, was used to fabricate the samples studied in this work. This technique is widely used in both the hard drive and semiconductor industries as it can cheaply and uniformly deposit a wide range of materials including conductive, insulating, magnetic and dielectric compounds and alloys [8].

The sputtering system used for the fabrication of the samples in the work was put together by AJA International (Fig. 3.1(a)). The main parts of the sputtering system are the main chamber, load lock, power supplies, and computer interface. Within the main chamber are seven sputtering sources, pictured in Fig. 3.1(b), which are commonly called 'guns'. Within the guns the target material to be sputtered is placed. The sputtering guns are arranged in a confocal pattern to ensure that each gun deposits uniformly over the entire surface area of the substrate. There are four power supplies which energize the sputtering guns: (3) radio-frequency (RF) sources and (1) DC source. Each power source may be operated simultaneously, which allows four guns to be on at one time. This is important to the samples in this work as it allows for co-sputtering, a method used to deposit alloy samples. To insert the substrates into the main chamber, they are first placed in the load lock and pumped down to \( \sim 10^{-6} - 10^{-7} \) Torr by a dedicated turbo pump. Once one of the substrate holders is
transferred manually into the main chamber, the gate valve between the main chamber and load lock is sealed and the sputtering process can begin. Much of the sputtering process that occurs in the main chamber can controlled remotely through Labview. This includes the temperature and pressure in the main chamber and the power to the sputtering guns. Figure 3.1(c) shows a high-temperature chromium deposition in action. A prominent feature in this photo is the localized plasma around the sputtering gun. The next section elucidates the sputtering process, which takes place within this glow discharge.
3.1.1 Magnetron Sputtering

The process of magnetron sputtering begins with a vacuum chamber pumped down to a low base pressure (better than $5 \times 10^{-7}$ Torr for the samples in this work) to reduce the amount of impurities. An inert gas like Argon, called the working gas, is introduced into the chamber. The working gas is ionized by a power source to create a plasma between the cathode (target material) and anode (substrate). The act of sputtering, which is the removal of surface atoms due to energetic particle bombardment, is just one of several processes that take place dynamically within the plasma. Figure 3.2 diagrams these processes: (1) Energetic free electrons, accelerating through the potential difference between the cathode and anode, collides with Ar atoms, knocking loose other electrons from the Ar outer shells. (2) The resulting Ar cation, attracted to the cathode, bombards and knocks loose some target material and more free electrons. Bits of the target material reach the substrate while some
of the free electrons become trapped in the field of the magnetron; both maintaining the plasma and feeding the formation of new Ar ions. (3) Finally, the Ar anion recombines with another free electron to become neutral Ar again.

The effect of the magnetron mentioned in step (2) can be seen visually as rings (or racetracks) on the sputtering targets in Fig. 3.1(b). Uneven target erosion by sputtering occurs within this track because ionization of the working gas is most intense above it. With more electrons confined to the magnetic field lines of the magnetron, the deposition rate on the substrate increases, but at the cost of inefficient use of the target. Once the racetrack depth nears the thickness of the target, it is unusable and must be replaced.

### 3.2 Structural Characterization

X-ray scattering is a non-destructive characterization method used to probe the structural details of a sample [9]. This technique is widely used in thin film magnetism as the magnetic properties of a sample are intimately related to its physical structure. X-ray scattering techniques reveal macroscopic information such as the phase of the material, crystal structure, film thickness and density. Microscopic details of the structure such as grain size and interfacial roughness can also be realized. To better understand the influence of these properties on the magnetism of the samples in this work, two methods of X-ray scattering were employed: X-ray diffraction (XRD) and X-ray reflectivity (XRR).

A Bruker AXS D8 Focus Diffractometer operating at characteristic wavelength of Cu Kα ($\lambda = 1.54056 \text{ Å}$) was used to perform the XRD and XRR measurements. A schematic of the experimental setup is shown in Fig. 3.3(a). In this figure, $\theta$ represents the incident angle of the x-ray beam, $\omega$ describes the angle between the sample and the incident beam, and $2\theta$ is defined as the diffracted angle. For the XRD scan, the diffractometer is operated in locked-coupled mode. In this mode, the
source is stationary and the sample and detector move in tandem to scan the desired
$2\theta$ range (typically 20°-80°) such that the incident angle $\theta$ is always equal to the half
of the diffracted angle, $2\theta$. This is also referred to as a symmetric scan. For the
XRR measurements, the diffractometer is operated in $2\theta$-$\omega$ mode, which is identical
to the locked-coupled mode in that it is also symmetric, except the scanning range is
smaller, between 0.1° and 6°.

3.2.1 Bragg’s Law

A model of the interference patterns of x-rays scattered by crystals was developed
by the father-son team, Sir W. H. Bragg and his son Sir W. L. Bragg. The Sirs
Bragg observed intensity peaks of scattered x-rays (called Bragg reflections) when
the following two conditions were true: (1) the incident angle of the x-rays was equal
to the scattering angle and (2) the path length difference of the x-rays was integer
multiples of their wavelength. Figure 3.3(b) illustrates how these two conditions form
a simple equation relating an integer multiple of the wavelength, $n\lambda$, and the angle
of the scattered x-ray, $\theta$, to the spacing between the planes of a lattice, $d$:

$$n\lambda = 2d\sin \theta \quad (6)$$

or

$$d = \frac{n\lambda}{2\sin \theta}. \quad (7)$$

When eq. 6 is satisfied, a diffraction pattern, or Bragg reflections, are observed.

3.2.2 X-Ray Diffraction

Consider the impact a large incident angle $\theta$ has on $d$ in equation 7 for $\lambda \sim 1$
Å. A large incident angle decreases the magnitude of $d$ to the order of atomic length
scales. Therefore observed Bragg reflections at wide angles are due to constructive
Figure 3.3: (a) A cartoon depiction of the diffractometer and scanning geometry used in XRD and XRR scans. (b) A schematic representation of Bragg’s law.

interference of the x-rays scattering from sets of periodic atomic lattice planes [10]. An analysis of the position and shape of the Bragg reflections as a function of $2\theta$ gives information about the sample’s crystal structure, orientation, texture (distribution of crystallographic orientations), and grain size. This is accomplished by fitting the peaks found in XRD scans to Gaussian or Lorentzian functions (or sometimes a combination of both). Two parameters of the fit are extracted: the peak center; to identify the peak, and the full-width-half-max (FWHM) of the peak; to estimate grain size.

Indexing the peaks in an XRD scan of sputtered films is relatively simple since the nominal structure of the film is known. For example, Fig. 3.4 presents an XRD scan of SiOx/Ta(60 nm). This scan was made in order to identify the phase and microscopic details of Ta that is sputtered at 60W DC power. Here, the peak was fit to a Lorentzian function. The center of the peak was found to be $x_c = 34.33^\circ$, corresponding to the (002) peak of the $\beta$-phase of Ta [11]. The FWHM of this peak, indicated by the arrows in the figure, was found to be $0.315^\circ$.

In an empirical equation called the Scherrer equation [12], the peak location
Figure 3.4: An XRD scan of SiOx/Ta(60nm), the red line is a fit to a Lorentzian function. The arrows denote the full-width half-maximum of the peak and the dashed line marks the peak center.

Figure 3.5: An XRR scan, with background subtracted, of SiOx/Ta(60nm). The red dots mark the index, n, of the Keissig oscillations. The inset shows the scan with background present, which is an artifact of the detector of the Bruker Diffractometer.
and width can be related to the grain size of the film by:

\[ L = \frac{K\lambda}{B(2\theta) \cos \theta}, \]  

(8)

where \( L \) is the estimated grain size, \( K \) is the Scherrer constant; a dimensionless shape factor, \( \lambda \) is the wavelength of the radiation, \( B \) is the FWHM of the peak, which should be reported in radians, and \( \theta \) is the Bragg angle. The grain size for SiOx/Ta(60 nm) was estimated to be 27 nm, which is in good agreement with previously reported grain size of sputtered \( \beta \)-Ta [13]. In a closing note, the Scherrer equation must be used thoughtfully as it assumes the only contributing factor to peak broadening is crystallite size.

### 3.2.3 X-Ray Reflectivity

In XRR scans, a small angle of incidence \( \theta \) is used to measure the Bragg reflections at \( 2\theta \). This small entrance angle has the effect of longer traveled paths for the x-rays and thus a sensitivity to periodicities of a larger length scale [10]. Figure 3.5 presents a typical reflectivity scan. The two main features of the scan are a plateau, which abruptly ends at a critical angle, \( \theta_c \), followed by several oscillations of intensity that decay at higher angles. The nearly constant reflectivity at small angles is due to total reflection of the x-ray beam. This is because the index of refraction of x-rays for almost all materials is less than one, so when the x-ray beam travels from air to a surface at an angle \( \theta < \theta_c \) total external reflection occurs (this is more apparent in the inset of Fig. 3.5). The critical angle can be related to the density of the film. The occurrence of intensity oscillations, called Kiessig oscillations, can be understood in the same way as Bragg’s law. In XRR, however, the intensity maxima occur due to constructive interference of the x-rays scattering from the differences in density of the substrate and film layer(s). This is how the XRR scan is used to determine the thickness of the film. Procedurally, this is accomplished by plotting the index, \( n \), of
each Kiessig oscillation (marked in Fig. 3.5) versus $\frac{2 \sin \theta_n}{\lambda}$. The scatter plot is fit to a line with the slope equal to $d$, the thickness of the film. This is how the sputtering rates for each target is determined.

### 3.3 Magneto-Optical Kerr Effect

The magneto-optical Kerr effect (MOKE) has wide applications in the study of magnetism due to a combination of straightforward implementation and versatility [14] [15] [16]. As an optical probe it provides non-destructive, localized magnetic information governed by the divergence of the light source beam, typically a laser, and the skin depth of the beam associated with the material under investigation. The material can be any magnetic material with a reflective surface; this includes metals, magnetic ceramics (ferrites), and magnetic semiconductors.

#### 3.3.1 Background

MOKE is a quantum mechanical effect and its physical origin lies in the spin-orbit interaction [14]. This can be understood by considering the magnetic moment of an electron in a material, $\mu$, and its motion, $v$, influenced by the electric field of incident light, $E$. In the rest frame of the electron, it feels a magnetic field $B \sim v \times E$. The spin-orbit coupling then has the form $\sim \mu \cdot (v \times E)$, which links the magnetic moment to the electron’s motion [17]. From the motion of the electron the probability current $j$ can be calculated and then averaged over the unit cell, and then the whole crystal. In doing so, the current density $J$ is found. It is well known that the current density induced by the incident radiation $E$ is related to the conductivity tensor $\hat{\sigma}$ through Ohm’s law $J = \hat{\sigma} \cdot E$, thus connecting the quantum mechanical phenomena of spin-orbit interaction with the (macroscopic) optical properties of a material governed by $\hat{\sigma}$ [18]. In metals, this is related to the permittivity tensor $\hat{\epsilon}$ by $\hat{\epsilon} = 1 + \frac{4\pi i}{\omega} \hat{\sigma}$

The phenomenon of MOKE is a rotation of the plane of polarization, elliptic-
ity, and/or reflectivity of linearly polarized light upon reflection from a magnetized surface. The changes which occur depend on the orientation of a sample’s magnetization vector, $\mathbf{M}$, with respect to the surface of the sample and to the optical plane of incidence. Here, the plane of incidence is defined as the plane spanned by the surface normal and wave propagation vector, $\mathbf{k}$. The polarization of light is described using the $\mathbf{p}$ and $\mathbf{s}$ vectors, denoting the components of the electric field vector parallel and perpendicular to the plane of incidence, respectively.

There are three geometries that categorize the Kerr effect: polar, longitudinal, and transverse. They can each be visualized in Figure 3.6 (a)-(c), respectively. In the polar geometry, the sample’s magnetization, $\mathbf{M}$, is out of the plane and also parallel to the plane of incidence. Light that is linearly polarized will experience a rotation of the plane of polarization and/or ellipticity. In the longitudinal and transverse geometries, $\mathbf{M}$ is in the plane of the sample but either parallel or perpendicular, respectively, to the plane of incidence. The effect on linearly polarized light in the longitudinal geometry is the same as the polar, either a rotation of the polarization or an ellipticity is introduced upon reflection. In the transverse geometry, however, the only effect on incident light is a change in reflectivity. It is important to note that for a sample with an arbitrary direction of $\mathbf{M}$, a combination of these three effects can exist and care must be taken to only measure one of them.

### 3.3.2 Development of a MOKE Magnetometer

**Optical Components.** The optical components of a MOKE magnetometer are very simple. As seen in Fig. 3.7, the foundation is a light source, two polarizers, and a photodetector. The light source chosen for our setup is a 5mW 655nm laser diode module (Edmund Optics). It has the ability of output modulation, a feature used to electronically chop the laser at a reference frequency for lock-in detection. The two polarizers are mounted linear polarizers (Thorlabs) with an extinction ratio.
Figure 3.6: Schematics of the (a) polar, (b) longitudinal, and (c) transverse MOKE geometries and the effect on arbitrarily polarized light incident at angle $\theta$. The dotted line represents the optical plane of incidence, the dashed line represents the surface normal and the primed components denote a change in amplitude upon reflection.

Figure 3.7: Principle components and geometry of the homebuilt MOKE. The laser, detector, and polarizers are mounted to an old table found on campus, cut into an L-shape, and fit between the pole faces of the electromagnet.
Figure 3.8: A block diagram of the components in our MOKE apparatus. The TTL output of the lock-in amplifier (LIA) modulates the intensity of the laser diode module which is phased-locked to the LIAs reference oscillator. The Hall probe (HP) and LIA are connected to digital multimeters (not pictured) which are interfaced to Labview.

of 100,000:1. Since most semiconductor lasers are linearly polarized (polarization ratio 400:1 for our laser), the purpose of the first polarizer is to improve the beam polarization ratio. The second polarizer serves as the analyzing polarizer to measure the Kerr rotation. The intensity of the beam is transduced into a voltage by an amplified Si photodetector (PDA36A, Thorlabs).

**Non-optical Components.** The MOKE apparatus was designed with a modular approach to add versatility to the setup. Interchangeable sample stages are the modular feature that allows for a quick change of measurement geometry without disturbing sensitive optical components. The sample stages are composed of heavy wood, aluminum, or microscope slides although the most useful sample stages are made from LEGOs. LEGOs are surprisingly ideal for magnetic environments. They are rigid; primarily right parallelepipeds, which is useful for quick alignment; and modular in nature. Many readily available parts have standard 45 degree angles and flat surfaces which are useful for mirror or sample mounting. It is essential that stages be built from a non-magnetic material that is sturdy and heavy to reduce noise due to mechanical vibrations, like from a magnet or large power supply. It is worthwhile
Figure 3.9: (a) A top-down view of the stage for the polar geometry. The stage orients the sample such that its out-of-plane magnetization is parallel to the applied field. Three Si mirrors reflect the incoming beam to be incident at near-normal angles and can be adjusted. (b) A picture of the rotational stage used for transverse MOKE measurements.

to construct spring clips or a similar mechanism that holds a sample down rather than to use tape, which has a long mechanical relaxation time that causes significant and uncontrollable signal drift.

Figure 3.9(a) is a photo of a sample stage made from LEGOs, which is used for measurements in the polar geometry. It has three small mirrors, two of which can rotate such that the angle of incidence on the sample can be varied. Figure 3.9(c) is a photo of a rotational sample stage which we use to study in-plane magnetic anisotropy in the transverse geometry. Its main feature is a single-axis rotating LEGO which allows the rotation a sample with respect to the applied field. This sample stage is used to study in-plane magnetic anisotropy.

**Data Acquisition and Analysis.** Labview is used to collect data from the
lock-in amplifier and Hall probe. The program displays the hysteresis loop (Fig. 3.10(a)) and a time-dependent signal from the lock-in (Fig. 3.10(b)) and outputs a data file containing the field measurement, lock-in signal measurement, and time. The time-dependent signal is used to identify and diagnose signal drift in the measurement. Signal drift usually originates from sample mounting. Another source of signal drift is due to temperature changes in the surrounding environment of the setup. A linear drift can correlate to temperature changes in the photodetector as dark current in the photodiode increases linearly with temperature. A typical temperature coefficient of a photodiode is 0.25% increase in current per degree C [19].

Samples are often encountered with low signal-to-noise ratios. To combat this, multiple MOKE measurements are taken (anywhere between 5-100 loops) and time-averaged. Figure 3.10(a) is an example of a transverse MOKE measurement on a SiOx/Ni(30nm) with 9 hysteresis loops. A Matlab program was written to average and analyze the MOKE data. An outline of the program is as follows: First, the time-dependent lock-in signal (Fig. 3.10(b)) is fitted to a polynomial. Then, the residual of the fit is used as the data to be time-averaged. The residual is plotted against the applied field in Fig. 3.10(c). The branches of each hysteresis are extracted and then interpolated such that each branch has the same number of data points. The increasing and decreasing branches are averaged separately. The two averages are then brought together as the final loop and is normalized (Fig. 3.10(e)). In addition, the program extracts the coercivity, $H_C$, exchange bias, $H_{EB}$ and squareness, $\frac{M_B}{M_S}$ of the averaged loop by recording the values of hysteresis as it crosses the x- and y-axis.
Figure 3.10: (a) Transverse MOKE measurements on a SiOx/Ni(30 nm) thin film. The time-dependent lock-in signal (b) shows a linear signal drift. The residual of the linear fit is plotted versus applied field (c) and time (d). The average of the fitted data is presented in (e) and normalized.
4 Results and Discussion

4.1 Induced Anisotropy in Permalloy

The Stoner-Wohlfarth model [20] is simulated in Matlab to determine the shape of the hysteresis loops of a uniaxial anisotropic material as a function of applied field strength and easy axis orientation. A common method for inducing a uniaxial anisotropy is to apply a magnetic field during the deposition of a magnetic thin film [21]. The effect of this deposition field on the magnetic anisotropy of a Permalloy thin film was investigated utilizing the MOKE apparatus in the transverse geometry. The simulation and experimental data are compared and deviations from the model are discussed.

4.1.1 Materials and Methods

Two samples with the nominal structure SiOx/Ta(50 Å)/Py(300 Å)/Ta(50 Å) were grown by magnetron sputtering at ambient temperature. The thin, amorphous Ta buffer was deposited to ensure a smooth substrate for the Py layer while preventing diffusion into the thermal oxide of the Si substrate. The Ta cap has the same purpose; to prevent oxidation upon removal from the sputtering system for magnetic characterization. The two films were deposited simultaneously using a custom substrate plate which delivers a shielded 250 Oe in-plane magnetic field on one side while the opposite side has a field below the detection limits of a calibrated Gaussmeter.

The magnetic measurements were carried out using the MOKE apparatus described in the previous section. The measurement geometry was transverse, since
Figure 4.1: The coordinate system for the Stoner-Wohlfarth model. The x-y plane is parallel to the plane of the sample. The sample is rotated by an angle $\phi$ to the applied field, $H$, which is fixed to the x-axis. The easy axis, denoted EA, is fixed to the sample while $M_s$ is free to rotate in the sample plane by an angle $\theta$ until an equilibrium angle $\theta'$ (not pictured) is reached.

Py is known to have in plane magnetization. The rotational LEGO sample stage was used to vary the orientation of the thin film with respect to the applied field. Several hysteresis loops were measured in steps of 10 degrees. The hysteresis loops were time-averaged using the Matlab program described in the previous section. The values of the squareness were also extracted using the Matlab program.

4.1.2 Simulation of Model

A widely used and highly instructive model of the magnetic behavior of a ferromagnet with uniaxial magnetic anisotropy (UMA) is the Stoner-Wohlfarth (SW) model [20] [22] [23]. It is the simplest analytical model which produces hysteresis. It assumes that the ferromagnet behaves as a single magnetic domain or that the domain walls are pinned and aligned with the easy axis. More importantly, it is assumed that the magnetization reversal mechanism of the system is coherent rotation of the magnetization, in other words, the magnetization remains uniform throughout the sample as its orientation changes.

A magnetic material has an energy function that describes the equilibrium
orientation of the magnetization as a function of external parameters. The model begins by constructing the total Gibbs free energy of the system, which is composed of two terms, the uniaxial magnetic anisotropy contribution from the deposition field and the Zeeman term. The energy function is as follows:

\[ F = K_1 \sin^2(\theta - \phi) - HM_s \cos(\theta), \quad (9) \]

where \( K_1 \) is the uniaxial anisotropy constant, \( H \), the applied field, and \( M_s \) is the saturation magnetization. The anisotropy axis (easy axis) is along the x-direction. The angle \( \theta \) follows the orientation of the magnetization with respect to the x-axis and \( \phi \) tracks the angle the easy axis makes with the x-axis (see Fig. 4.1).

By defining the reduced field, \( h = \frac{M_s}{2K_1} H \), equation (9) is now:

\[ f = \frac{1}{2} \sin^2(\theta - \phi) - h \cos(\theta). \quad (10) \]

This makes the problem dimensionless and easier to solve numerically. For certain values of \( \phi \) (namely, 0, \( \frac{\pi}{4} \), and \( \frac{\pi}{2} \)), however, equation (10) can be solved analytically [23] via optimization. Doing so for the hard-axis case of \( \phi = \frac{\pi}{2} \) reveals a way to experimentally determine the scaling factor \( \frac{M_s}{2K_1} \).

The energy density at \( \phi = \frac{\pi}{2} \) is a minimum when

\[ \frac{\partial f(\theta, \phi, h)}{\partial \theta} = 0 = -\cos \theta \sin \theta + h \sin \theta, \quad (11) \]

or, when

\[ \sin(\theta) = 0 \quad (12a) \]
\[ \cos(\theta) = h, \quad (12b) \]
and when the second derivative is positive:

\[
\frac{\partial^2 f(\theta, \phi, h)}{\partial \theta^2} = -\cos^2(\theta) + h\cos(\theta) > 0.
\] (13)

Equation 12(a) has the solutions \( \theta = 0, \pi, \ldots \), which is recognized as the saturated states, i.e. the magnetization vector pointing parallel and antiparallel, respectively, to the applied field direction. This is confirmed by the stability condition, eq.13, which states that \( \theta = 0 \) is stable only for \( h > 1 \) and \( \theta = \pi \) is stable for \( h < -1 \).

But what about the intermediate region \(-1 < h < 1\)? First, it is instructive to make the substitution \( \frac{M}{M_s} = \cos(\theta) \). This is just the projection of the magnetization vector along the field direction. Plugging that in to Eq. 7b yields

\[
\frac{M}{M_s} = h,
\] (14)

which is the equation of motion for the magnetization in fields below saturation \((-1 < h = \frac{M}{2K_1}H < 1\)). This equation is the general equation for the hard-axis magnetization process of a uniaxial material. It is linear for fields up until \( h = \pm 1 \). Therefore, the slope of the hard-axis magnetization curve at fields between positive and negative saturation is equal to \( \frac{M_s}{2K_1} \).

Figure 4.2 shows a MOKE hard axis loop of the SiOx/Ta(50 Å)/Py(300 Å)/ (Ta50 Å) sample that was sputter-deposited under the influence of an applied magnetic field. The low field data between positive and negative saturation was fit to a line in the inset of Fig. 4.2, with \( \frac{M_s}{2K_1} = 0.151 \text{ Oe}^{-1} \).

With the scaling factor known, a magnetization curve can now be simulated in Matlab by finding the equilibrium magnetization angle \( \theta' \) which satisfies Eq. 11 for various values of \( \phi \) and \( h \). We can again relate the equilibrium angle \( \theta' \) to the magnetization by using the relationship \( \frac{M}{M_s} = \cos(\theta') \). A simple Matlab program was written to solve (3). A copy of the code is given in the appendix. An outline of the
Figure 4.2: Transverse MOKE loop of SiOx/Ta(50 Å)/Py(300 Å)/Ta(50 Å) at \( \phi = 90 \) deg. The red line is a guide to the eye. The inset is a zoom of the linear fit between saturation with the slope \( M_s/2K_u = 0.151\text{Oe}^{-1} \).

The code follows:

1. Two arrays are defined for various values of \( \phi \) and \( H \); for example: \( \phi = 0 : \pi/180 : 2\pi \) and \( H = -40 : 1 : 40 \) Oe; the latter determined by the field range used in the experimental data

2. A user-defined function is created for Eq.10 with the scaling factor included and the variables \( \theta, \phi, \) and \( H \)

3. Matlab’s \texttt{fminsearch} function finds the minimum of Eq.10 within two nested for loops that cycle through the various values of \( \phi \) and \( H \)

4. The output of \texttt{fminsearch} is a \texttt{length}(\( H \)) by \texttt{length}(\( \phi \)) matrix of equilibrium magnetization angles \( \theta' \), where the columns of the matrix are the values of \( \theta'(H) \) at each angle \( \phi \)

5. The cosine of each element in the matrix is taken to relate the normalized magnetization to the equilibrium angle, i.e. \( m = \frac{M}{M_s} = \cos\theta' \)
6. The program writes two text files: one with the column of $H$ values concatenated with the matrix of $m$. The second is the column of $\phi$ values concatenated with the transpose of matrix $m$.

The first file is used to plot the hysteresis curves $m(h)$ for the different angles $\phi$. The second file is for plotting certain characteristics of the hysteresis as a function of angle from the easy axis, i.e., for plotting the angular dependence of the squareness ($\frac{M(H=0)}{M_S}$). In the next section, the simulation results for the SW model will be compared to the transverse MOKE data from a field deposited sample of SiOx/Ta(50 Å)/Py(300 Å)/Ta(50 Å).

4.1.3 Application of Model

The $m(H)$ loops from the simulation for several $\phi$ have been plotted in Fig. 4.3 on top of the MOKE data for the field deposited sample. There is excellent qualitative agreement for almost all the orientations of $\phi$. The shape of the loops are captured by the Stoner-Wohlfarth model with the exception of $\phi$ along the hard axis (Fig. 4.3(e)). The non-zero remanence in this case can be plausibly explained by the existence of an anisotropy axis distribution in the film, which is a natural occurrence in polycrystalline materials [24] [25]. Such a distribution can be incorporated into more detailed models.

For most orientations of $\phi$ there is good quantitative agreement between the model and experiment. However, for $\phi \pm 10^\circ$ from the easy axis (Fig. 4.3(a)-(c) and (g)-(i)), there is a clear overestimate of the coercive field. This is can be explained by the assumption in the model that the reversal mechanism is coherent rotation for all orientations of $\phi$. Magnetization reversal by coherent rotation is not always the most energetically favorable, especially for orientations of $\phi$ near the easy axis [26]. This is especially true for Permalloy, which has very little magnetocrystalline anisotropy and therefore the reversal processes are dominated by magnetostatic interaction [27].
Other modes of reversal exist which are mediated through domain walls motion. At a thickness of 300 Å, the sample is predicted to have Néel domain walls [23]. This type of domain wall forms in thin films (thickness< 400 Å) when the magnetostatic energy can be reduced in the reversal process by the spins rotating 180° in the plane of the film, as opposed to rotating in the plane normal to the surface (Bloch walls). Further, Permalloy is also known to form cross-tie domain wall structures at this thickness when the angle $\phi$ near the easy axis [28] [29]. This type of domain wall forms to reduce the magnetostatic energy by flux closure; two 90° domain walls of opposite directions form at the ends of two antiparallel Néel walls to reduce the free poles at the ends of the two Néel domains. At any rate, domain wall motion during the magnetization reversal costs less energy than uniform rotation of the magnetization, thus the coercive field is reduced at $\phi$ near the easy axis when this type of reversal mechanism dominates.

The angular evolution of the remanence provides a picture of the symmetry of the magnetic anisotropy in the sample. Figure 4.4 plots the squareness (remanence normalized by $M_S$) from MOKE as a polar plot. The shape is easily identified as being uniaxial in symmetry. At $\phi=0, 180^\circ$ the easy axis the squareness is a maximum and along the hard axis the squareness approaches zero. The result from the simulation fits well to the data. The deposition field induced a clear uniaxial symmetry in the sample although there exists a distribution of easy axes, which is evident by the non-zero remanence and expected for a polycrystalline film.

4.1.4 Conclusions

A very simple model was used to analyze the magnetic behavior of a Py thin film with induced uniaxial anisotropy. The model was compared to experimental MOKE data and deviation from the model was discussed in terms of the assumptions made in the model. The absence of domain wall effects in the model was the source of the
Figure 4.3: Hysteresis loops from the transverse MOKE measurements as a function of $\phi$. The red line is the result of the simulation at the same angle $\phi$. 
Figure 4.4: The angular dependence of the squareness ($\frac{M_r}{M_s}$) of SiOx/Ta(50 Å)/Py(300 Å)/Ta(50 Å) along with transverse MOKE loops at various angles. The red line is the result of the model.
overestimation of the coercive fields for the experimental data at φ near the easy axis. A non-zero remanence at the hard axis of the experimental data was also not predicted by the model due to an existence of a distribution of easy axes in the polycrystalline film. Regardless of the broad assumptions made, important and instructive aspects of the basic physics of magnetic reversal phenomena can be made by implementation of the Stoner-Wohlfarth model.

4.2 Exchange Bias of Mu-metal

Contents of this chapter have been published in Ref. [30].

The focus of this work was on inducing exchange bias [5] [31] in Ni$_{77}$Fe$_{14}$Cu$_5$Mo$_4$, which is also known as mu-metal. Mu-metal is a member of the Permalloy family of alloys and is well known for its large permeability and saturation magnetization. It also has nearly zero magnetostriction and a very small magnetocrystalline anisotropy [32] [33]. Inducing a unidirectional anisotropy in soft ferromagnets such as mu-metal is useful for introducing additional control over devices such as magnetic sensors which utilize the giant magneto-impedance effect (GMI) [34]. Bulk mu-metal has been shown to have a large GMI ratio (300%) and also a high sensitivity (20%/Oe) [35] [36], but its exchange bias properties had not yet been reported.

4.2.1 Materials and Methods

The structural and magnetic properties of several sets of Ni$_{77}$Fe$_{14}$Cu$_5$Mo$_4$/Fe$_5$0Mn$_{50}$ (NiFeCuMo/FeMn) with varying ferromagnet thickness. Figure 4.5 presents a schematic representation of all the samples fabricated for this study. The bilayer was deposited on different buffer layer compositions. The bilayer and buffers were deposited on 140nm thick thermal oxide of Si wafers (amorphous), denoted by SiOx. The buffer layers chosen were 50 Å of Ta and 300 or 800 Å of Cu. The thickness of the antiferromagnetic material, FeMn, was kept at a constant 150 Å for all the bilayers. This
thickness was chosen so that the blocking temperature (400K) was independent of
the antiferromagnet thickness [37].

The preparation of the substrates for sputtering includes ultrasonically cleaning
them in acetone for 5 min, followed by methanol for 5 min. The substrates are
then blown dry with Nitrogen gas before being inserted into the load lock. The sam-
pies were grown at ambient temperature in 3mtorr of ultra high purity Argon gas.
The base pressure for the fabrication of all samples was better than 20 nTorr. The
compositions of NiFeCuMo and FeMn were the same as the sputtering targets, i.e.
there was no cosputtering. The targets were presputtered for 10 min prior to each
deposition.

4.2.2 Structural Analysis

X-ray diffraction results for Cu k$_{\alpha}$ radiation are shown in Fig. 4.6 for each buffer
layer type. The thicknesses of the NiFeCuMo and FeMn are 400 Å and 150 Å, respec-
tively. The (111) orientation is specifically of interest because it is known to yield
the largest exchange bias when using FeMn as the antiferromagnet [38]. Each sample
shows shifted (111) peaks relative to the bulk. For the Ta buffered samples, NiFe-
CuMo shows the least strain along the grown direction relative to the bulk value. In
addition, the NiFeCuMo and FeMn peaks overlap, which was confirmed by sequential
XRD after deposition of each individual layer (not pictured). The 300 ÅCu-buffered
samples have the weakest (111) texturing out of all the samples studied. This indi-
cates that this thickness of buffer has low structural quality.

The 800 Å-Cu buffered samples show a more prominent (111) texturing in Cu than the 300 Å-Cu buffer, however, the (111) texture in the NiFeCuMo/FeMn is not strong in both. While the Ta-buffered samples show the most coherent (111) texturing of NiFeCuMo/FeMn, a simple Scherrer analysis indicates that the coherence length is only about 80 Å. The inset of Fig. 4.6 shows the crystallinity of 1000 Å of Cu is improved over the 800 Å of Cu, but the large breadth of the peak indicates there is still room for improvement.

Figure 4.6: XRD scans of each buffer type deposited before the NiFeCuMo(400 Å)/FeMn bilayer. The inset shows the crystallinity of 1000 Å of Cu.

4.2.3 Magnetic Analysis

Two control samples of Cu(300 Å)/NiFeCuMo(200 Å)/Cu(300 Å) were deposited simultaneously using a custom substrate plate [39] which delivers a shielded 250Oe
in-plane magnetic field on one side while the opposite side has a field below the detection limits of a calibrated Gaussmeter. Since the exchange bias is set with an applied field during deposition, the purpose of these control samples was to confirm that the field during deposition had no effect on the deposition rate. XRR scans were used to confirm this. VSM was used to measure the hysteresis loops of the two samples. As shown in Fig. 4.7 (a), the sample deposited in no field is magnetically isotropic, with no difference in hysteresis loop shape for the magnetization measured along two orthogonal directions. In contrast, Fig. 4.7(b), the field-grown sample, displays uniaxial magnetic anisotropy with the easy axis corresponding to the deposition field axis. The coercivity slightly increases along the easy axis, while the hard axis coercivity has not measurably changed relative to the zero-field sample. The saturation fields are in line with previous results on NiFeCuMo thin films.

Relative to the control samples, a clear exchange bias develops when FeMn is deposited, in the presence of a magnetic field, onto NiFeCuMo. Figure 4.8 dis-
Figure 4.8: MOKE hysteresis loops measured along the easy axis of the three different buffer layer types. The black loop is SiOx/Cu(300 Å)/NiFeCuMo(200 Å)/FeMn(150 Å), the blue loop is SiOx/Cu(800 Å)/NiFeCuMo(200 Å)/FeMn(150 Å) and the red loop is SiOx/Ta(50 Å)/NiFeCuMo(200 Å)/FeMn(150 Å).

plays room temperature hysteresis loops measured along the easy axis for the NiFe-CuMo(200 Å)/FeMn(150 Å) bilayers deposited on the three different buffer layer compositions. The exchange bias field, $H_{EB}$, of the Cu(800 Å)-buffered (blue triangles) sample is nearly double that of the Ta(50 Å) buffer (red circles). The coercivity, $H_C$, is lower in the Ta(50 Å) buffered samples than the Cu(800 Å) buffer. In contrast, the samples with a Cu(300 Å) (black squares) buffer have both the highest $H_{EB}$ and $H_C$ of the lot, which is likely due to the ill-defined structure of the Cu(300 Å) buffer apparent in the XRD scans. The ill-defined structure may result in more pinned uncompensated moments per unit area at the NiFeCuMo/FeMn interface, which is known to increase the exchange bias [40]. Additionally, the Cu(300 Å) buffered samples have an enhanced switching field distribution which indicates that the soft magnetic properties of the mu-metal were essentially lost.
Figure 4.9: (a) The $\cos \theta$ dependence of the exchange bias field for Ta(50 Å)/Mu(200 Å)/FeMn(150 Å). (b) A global view of the exchange bias and coercive fields which are inversely proportional to the ferromagnet thickness. The thin lines are linear fits.

Figure 4.9(b) provides the thickness dependence of $H_{EB}$ and $H_C$ for each sample grown on SiOx. To determine the exchange bias field for any one sample, hysteresis loops were measured as a function of angle with respect to the deposition field and fit to a cosine function. Each sample displayed an $H_{EB} \cos(\theta)$ dependence, as shown in Fig. 4.9(a), with the amplitude taken to be the exchange bias field for the sample. Both $H_{EB}$ and $H_C$ were found to be inversely proportional to the NiFeCuMo thickness, which is expected since exchange bias is an interface effect.

Using the values of $H_{EB}$ measured along the easy axis direction, we can determine the interfacial energy per unit area according to:

$$H_{EB} = \frac{J_{int}}{M_S t_{FM}},$$

(15)

where $M_S$ and $t_{FM}$ are the saturation magnetization and thickness of the NiFeCuMo, respectively. $M_S$ was measured by VSM and found to be 265 emu/cm$^3$ and independent of the thicknesses studied. Figure 4.10 shows that linear fits of the exchange bias as a function of $1/M_S t_{FM}$ yield $J_{int} = -11.7 \pm 1.3$ marg/cm$^2$ for the Ta buffer,
Figure 4.10: The interfacial energy per unit area $J_{int}$ is the slope of the linear fits of $H_{EB}$ vs $1/M_{S}\tau_{FM}$.

-82.2 ± 2.1 merg/cm$^2$ for 300Å-Cu buffer, and -37.1 ± 5.1 merg/cm$^2$ for the 800Å-Cu buffered bilayers. Despite the seeming large spread, each $J_{int}$ value is in agreement with previous energy densities reported using FeMn(111) as the antiferromagnet [5].

4.2.4 Conclusions

Together, these results show that mu-metal exhibits classic exchange bias behavior when grown in contact with FeMn. The differences in magnetic properties between the Ta-buffered and Cu-buffered NiFeCuMo/FeMn bilayers are significant with respect to potential application in devices such as low-field sensing. The origin of the differences appear to be structural in nature. While annealing or higher temperature deposition may improve the structure, including possibly reducing the required Cu buffer thickness, this may come at the price of interdiffusion and subsequent loss of
mu-metals valuable soft magnetism. Note that NiFeCuMo films may be susceptible to deposition-induced structural perturbations: we find it necessary to rotate the samples during growth in order to obtain reproducible magnetic properties; growing with the sputtering flux at a fixed angle relative to a stationary substrate leads to unexpected (and difficult to control) magnetocrystalline anisotropy. It is possible that this structural sensitivity is playing a significant role in response to the differences in strain induced by the amorphous Ta and polycrystalline Cu buffers.

Although both buffer materials allowed (111) texturing of NiFeCuMo and FeMn, the samples with the Ta buffers preserved the soft magnetic properties of the mu-metal most effectively. The 300 Å-Cu buffer had low quality (111) texturing, which led to significantly enhanced exchange bias and coercive fields, along with a broadened switching field distribution. One notable result here is the ability to preserve the soft features the mu-metal while inducing a unidirectional anisotropy when using Ta as the buffer. From a practical spintronics point, Ta may also be the most beneficial substrate because the relatively thin layer will lead to greater current density in the magnetic layers. This may thus impact devices and structures employing soft magnetic materials, such as giant magnetoimpedance and related sensors.
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About the Author

Tatiana Marie Eggers was born in Houston, Texas on Oct. 19, 1984 to parents Tania and Frank Eggers. Tatiana studied mathematics and physics as an undergraduate student at the University of South Florida. She graduated with a B.S. in physics in 2012. She began research as an undergraduate student under the advisement of Casey W. Miller in the Spintronics Lab group and immediately began work using the Magneto-Optical Kerr Effect to characterize magnetic films. She continued her work with the group as a graduate student and went on to become proficient in fabricating thin films by magnetron sputtering.

Apart from research, Tatiana enjoys reading post-apocalyptic novels, playing video games, and watching Star Trek films and TV episodes over and over. She feels that Star Trek played a major role in her decision to pursue an advanced degree in physics.