

Perpendicular Magnetic Anisotropy in Ion Beam Sputtered Co/Ni Multilayers

By Boris Rasin

Submitted to the Department of Materials
Science and Engineering in Partial
Fulfillment of the Requirements for the
Degree of

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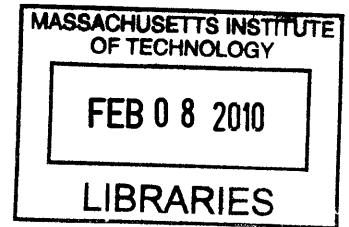
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Signature of Author.....
Department of Materials Science and Engineering

Certified by
Caroline A. Ross
Toyota Professor of Materials Science and Engineering
Thesis Supervisor

Accepted by
Lionel C. Kimerling
Professor of Materials Science and Engineering
Chair, Undergraduate Committee

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Abstract

Co/Ni multilayers display perpendicular magnetic anisotropy and have applications in magnetic devices that could lead to a large increase in the density of magnetic storage. Co/Ni $10 \cdot (2 \text{ \AA} \text{ Co} / 8 \text{ \AA} \text{ Ni})$ and $10 \cdot (2 \text{ \AA} \text{ Co} / 4 \text{ \AA} \text{ Ni})$ multilayers were deposited with ion beam sputtering on either ion beam sputtered copper or direct current magnetron sputtered gold buffer layers of various thicknesses. The effect of the roughness and the degree of (1 1 1) texture of the buffer layers and the multilayers on the perpendicular magnetic anisotropy of the deposited multilayers was examined. In addition the effect of the deposition method used to fabricate the samples, ion beam sputtering, was analyzed. The magnetic behavior of the multilayers was examined with alternating gradient magnetometry and vibrating sample magnetometry, the structure of the buffer layers and the multilayers was characterized with X-ray diffraction, and the roughness of the surface of the multilayers was characterized with atomic force microscopy. None of the deposited films showed perpendicular magnetic anisotropy and instead showed parallel magnetic anisotropy which was found to have occurred for every sample due to either a low degree of (1 1 1) texture in the buffer layer and the Co/Ni multilayer, a too high degree of roughness in the buffer layer and the Co/Ni multilayer or a combination of these two factors. In addition it was hypothesized that as the samples were deposited with sputtering, diffusion and alloying at the multilayer interfaces may have contributed to the multilayers having parallel magnetic anisotropy instead of perpendicular magnetic anisotropy.

Thesis Supervisor: Caroline A. Ross

Title: Toyota Professor of Materials Science and Engineering

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Chapter 1: Motivation and Introduction

Perpendicular magnetic anisotropy in a thin film is the property of the thin film to have its easy magnetization direction, the magnetization direction where the internal energy of the film is at a minimum, perpendicular to the plane of the thin film.¹ Perpendicular magnetic anisotropy only occurs in specific thin film multilayers. One reason thin film multilayers with perpendicular magnetic anisotropy are of great interest is due to their applications in magnetic memory devices where bits can be encoded by the application of a current.^{2 3 4 5 6} These systems are of great interest as the current method of encoding bits in magnetic memory devices involves generating a magnetic field from a wire to reverse the magnetization direction of a magnetic device.^{5 6} This is a barrier to high density memory as the current required to generate a sufficient field to reverse the magnetization direction of a very small magnetic device is too high and the magnitude of the magnetic field generated from a wire decreases slowly with distance and can affect the magnetization direction of adjacent magnetic devices.^{5 6} Two memory devices where bits can be encoded by current and to which thin film multilayers with perpendicular magnetic anisotropy have applications to are thin film multilayer nanostructures, such as nanopillars, where bits are encoded by reversing the magnetization direction of the free layer of the nanostructure by spin torque from spin polarized current and systems where bits are encoded by current induced domain wall motion, domain wall switching.^{2 4 5} Thin film multilayers with perpendicular magnetic anisotropy are important for the former device as the utilization of a reference layer with perpendicular magnetic anisotropy and a free layer that either has perpendicular magnetic anisotropy or parallel magnetic anisotropy in these systems allows for magnetization reversal to occur at lower currents and at faster rates.² Thin film multilayers with perpendicular magnetic anisotropy are important for the latter system as it has been shown that in nanowires with perpendicular magnetic anisotropy the domain walls are more stable and a lower current is required to move these domain walls as compared to nanowires with parallel magnetic anisotropy.⁴ Thus, the development of thin films with perpendicular magnetic anisotropy is critical for the development of memory where bits can be written with current and is thus important for the development of fast and high density memory in the future.

Perpendicular magnetic anisotropy has been achieved in a variety of thin film multilayers such as Ni/Pt, Co/Pt, Co/Pd, Co/Ir, Co/Au, Nd/Fe and Pr/Fe.^{7 8 9} Of these systems Co/Pt has been the most widely studied system both fundamentally and in devices. One thin film multilayer with perpendicular magnetic anisotropy which has not yet been studied extensively is Co/Ni.¹⁰ Co/Ni multilayers were found to have perpendicular magnetic anisotropy in 1991 and since then there have been relatively few studies of this system.¹⁰ However, there has recently been renewed interest in the system for a number of nanomagnetic devices.^{2 4} One example is the reversal of the magnetization of nanopillars with current via spin torque recently presented by Mangin et al.² Magin et al. revealed that when Co/Ni was used in the free and reference layers of the nanopillars the current density required to reverse the magnetization of the free layer was

between 3 and 4 times smaller than when Co/Pt was used and the GMR ratio of the nanopillars with Co/Ni was 1.0% while the GMR ratio of the nanopillars with Co/Pt was .4%.²

Thus, we chose to fabricate various thin film Co/Ni multilayers with ion beam sputtering to characterize their behavior and to potentially use them for device fabrication in the future. Co/Ni multilayers with perpendicular magnetic anisotropy typically consist of a buffer layer with 4Å-20Å thin Co/Ni bilayers deposited on top where the thickness of the Co is typically around 2Å and the thickness of the Ni is typically 4Å to 8Å. In this study Co/Ni bilayers were deposited with ion beam sputtering on gold and copper substrates of various thicknesses. The structure of the multilayers and the buffer layers was characterized with X-ray diffraction and the surface was analyzed with atomic force microscopy. The magnetic behavior was analyzed with vibrating sample magnetometry and alternating gradient magnetometry.

Chapter 2: Literature Review

Origins of Perpendicular Magnetic Anisotropy in Co/Ni Multilayers

Daalderop et al. have proposed that the perpendicular magnetic anisotropy in Co/Ni multilayers, with the proper thickness for the cobalt and nickel layers and deposited under the proper conditions as will be discussed later, is the result of two aspects of the system.¹⁰ The first aspect of the system that Daalderop et al. propose is responsible for the perpendicular magnetic anisotropy is the magnetic interface anisotropy at the interfaces between the cobalt and nickel layers.¹⁰ According to Neel's model of surface anisotropy the magnetic interface anisotropy arises due to a decrease in symmetry at the interfaces.^{10 11} In turn the reduction of symmetry at the interfaces is a result of the atoms of each material being coordinated differently at the surface of each layer as compared to their coordination inside the layer.¹¹

A second aspect of the system that Daalderop et al. suggest is responsible for the perpendicular magnetic anisotropy is the electronic structure at the Co/Ni interfaces.¹⁰ More specifically Daalderop et al. suggest that perpendicular anisotropy occurs in Co/Ni multilayers as there is an appropriate number of valence electrons for the Fermi level to be near specific states.^{10 12} These specific states near the Fermi level have spin orbit interactions which cause perpendicular magnetic anisotropy.^{10 12}

Broeder et al. and Johnson et al. propose a third cause of perpendicular magnetic anisotropy in Co/Ni films.^{12 13} Broeder et al. propose that magneto-elastic anisotropy contributes to the perpendicular magnetic anisotropy.¹³ Magneto-elastic anisotropy is magnetic anisotropy that results from the straining of the crystal lattice.¹⁴ Broeder et al. propose that this magneto-elastic anisotropy in the Co/Ni multilayers has two sources.¹³ The first source is proposed to be strain resulting from defects formed in the layer during deposition.¹³ Examples of such defects are vacancies and dislocations.¹³ The second source is proposed to be the strain on the multilayer due to the buffer layer on which it is deposited.¹³ This strain occurs as if the buffer layer is coherent with the multilayer, meaning that the lattice planes of the buffer layer are continuous with the lattice planes of the Co/Ni multilayer across the interface between the two, then a strain is applied to the multilayer lattice.^{13 15 16} Further, as the number of layers in the multilayer increases the strain due to coherence leads to the formation of dislocations which also contribute to the magnetoelastic anisotropy.^{12 13} It is of interest to note that only certain buffer layers will have a lattice that matches that of the multilayer closely enough for a coherent interface to form between the buffer layer and multilayer.¹³ It also important to note that the very slight lattice mismatch between the cobalt and nickel layers also provides magnetoelastic anisotropy but it is very slight and favors in plane magnetization.^{10 13}

Of these three contributions to the perpendicular magnetic anisotropy the first two, the magnetic interface anisotropy and the electronic structure at the interfaces are the dominant contributions with magneto-elastic anisotropy playing a minor role.¹³

Phenomenological Model of Magnetic Anisotropy

For a Co/Ni multilayer magnetic anisotropy can primarily be modeled with the relation,

$$Kt = K_V t + 2K_S \quad (1)$$

where K is the total magnetic anisotropy per unit volume, K_V is the volume anisotropy of the Co and Ni, K_S is the interface anisotropy for the Co/Ni interface, and t is the thickness of a bilayer.¹⁷ We can define this equation more specifically for the Co/Ni bilayer systems if we expand the $K_V t$ term to get the relation,

$$Kt = K_V^{Co} t_{Co} + K_V^{Ni} t_{Ni} + 2K_S \quad (2)$$

Where, K , t , and K_S are defined the same as in Equation 1 and K_V^{Ni} refers to the volume anisotropy of the Nickel layer, K_V^{Co} refers to the volume anisotropy of the Cobalt layer, t_{Co} refers to the thickness of the cobalt layer and t_{Ni} refers to the thickness of the nickel layer.¹⁸ A positive value of K indicates perpendicular magnetic anisotropy and a negative value of K indicates parallel magnetic anisotropy.¹² A more positive value of K indicates stronger perpendicular magnetic anisotropy and more negative value of K means stronger in plane anisotropy.

The equation indicates all of the contributions to the magnetic anisotropy, namely the volume and interface anisotropies. In Co/Ni multilayers the volume anisotropy K_V is usually negative pushing the system towards parallel magnetic anisotropy and when perpendicular magnetic anisotropy is observed K_S is positive enough to overcome the negative K_V such that K comes out positive in equation 2.^{12 17 18} This is in line with the theory we just discussed as the interface anisotropy at the interfaces of Co/Ni layers causes perpendicular anisotropy in Co/Ni multilayers.¹⁰ It is important to note that volume anisotropy can include the magnetocrystalline anisotropy, the shape anisotropy and K_σ the magnetoelastic anisotropy.¹⁷ This later component can be positive and thus contribute to the perpendicular magnetic anisotropy.^{12 13} This also goes with the theory we discussed as magnetoelastic anisotropy can contribute to perpendicular magnetic anisotropy.^{12 13}

It is also important to note that Equation 1 can be used to determine K_S and K_V . For example if the ratio of t_{Co} to t_{Ni} in the bilayers is kept constant and the total thickness of each bilayer is increased then according to Equation 1 Kt should change linearly with t , the total bilayer thickness.¹⁰ Hence according to Equation 1 the slope of a line through a set of Kt for various t data where the ratio of t_{Co} to t_{Ni} is kept constant is K_V and the K intercept is $2K_S$.^{10 12} The use of this method for actual data can be seen in Figure 1 taken from Daalderop et al, where they kept the t_{Co} to t_{Ni} ratio constant in a Co/Ni multilayer as they increased the thickness and used the K intercept to find $K_S = .3 \text{ mJ/m}^2$ and the slope to find $K_V = -.39 \text{ MJ/m}^3$.¹⁰ (Figure 1) (Figure 1 from Daalderop et. al¹⁰) Similar approaches with Equation 2 can be used to find K_V^{Co} and K_V^{Ni} .^{12 18}

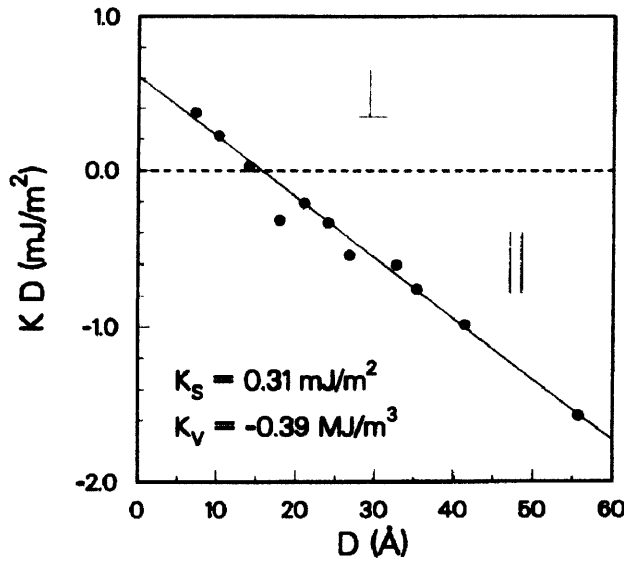


Figure 1: A plot of the total anisotropy constant per unit volume, K , multiplied by the bilayer thickness against the total bilayer thickness for a Co/Ni multilayer with the ratio of the Co thickness to the Ni thickness at 1:2.2. The intercept was used to determine the interface anisotropy, K_S , as the intercept is equal to $2 K_S$ and the slope was used to determine the volume anisotropy K_V as K_V is equal to the slope. (Figure from Daalderop et al.¹⁰)

Co and Ni Layer Thickness

The thickness of the cobalt and the nickel layers determines whether or not the system will display perpendicular magnetic anisotropy.¹⁰ In the case of Co/Ni multilayers the thickness of each layer affects the magnetic anisotropy of the system.¹⁰

Now we will discuss trends in the effect of the thickness of the Co and Ni layers on K . In line with our phenomenological model for K stated in Equation 1 and in Equation 2 as the thickness of the Ni layer, Co layer, or both Ni and Co layers if the ratio is kept constant increases, K changes linearly with the variable being increased.¹⁰ As in Co/Ni multilayers K_V is always negative K always decreases linearly with the increasing thickness of the Ni layer, Co layer, or both Ni and Co layers if the ratio is kept constant with a slope of K_V^{Ni} , K_V^{Co} or K_V respectively.^{10 12 13 18} This has been confirmed in numerous papers. Thus, as the thickness of one type of layer is increased K decreases.^{12 13 18} Alternatively if the thickness of both layers increases with the ratio of their thicknesses kept constant K decreases as well.¹⁰ From this point forward in this section we will discuss what occurs as the thickness of one layer increases to make explanation easier. However, these trends also apply when the thickness of both layers is increased with the ratio of the layers kept constant.

Thus, there are two scenarios we should consider. The first scenario is if a multilayer displaying perpendicular magnetic anisotropy. In this situation at small thickness of a given layer K will be positive and as the thickness of the layer increases K decreases linearly until it becomes negative and continues to become more and more negative as the thickness of the layer increases.^{12 13 18} Hence the material will have perpendicular magnetic anisotropy at small

thicknesses, when K is positive, which will decrease with the increasing thickness of a layer until in plane anisotropy will occur when the layer is sufficiently thick and K is negative.^{12 13 18} The magnetic anisotropy will then continue to be more and more in plane as the thickness of the layer increases and K becomes more and more negative.^{12 13 18} This scenario can be seen for a Co/Ni multilayer where the thickness of Ni is increased and the thickness of the cobalt is kept constant at 2Å from work done by Zhang et al.¹⁸ (Figure 2 from Zhang et al.¹⁸) It is of interest to note that Kt does decrease again when the thickness is very small out of line with the model.¹⁸(Figure 2) In another paper this decrease also occurs but at much smaller thicknesses such as less than 3Å and more gradually, possibly as the layer becomes too thin to be continuous when thicknesses of near one monolayer are reached.^{12 17} The second scenario we will consider is when the multilayer shows parallel magnetic anisotropy for all thicknesses of a given layer. In this case the Kt against layer thickness relationship should be exactly the same as for the first scenario except that Kt would be negative even at the smallest thicknesses. This can be visualized as shifting the Kt -Ni thickness line down. In this scenario K is always negative so there is no perpendicular anisotropy and the magnetic anisotropy becomes more and more in plane as the thickness of a given layer increases. In practice what occurs as observed in literature is that systems that only show in plane anisotropy have trendlines that suggest that at a very small layer thicknesses K will be positive, however, these thicknesses are often very thin and are smaller than the thickness of a monolayer which yields discontinuous films with low perpendicular magnetic anisotropy and data points for these thicknesses are not given even if they are larger than a monolayer.^{12 17} An example of such data can be found in Figure 3. (Figure 3 from Johnson et al.¹²)

Many thicknesses have been utilized for Co/Ni multilayers in literature. Two common thicknesses used are 2Å Co /4Å Ni and 2Å Co/ 8Å Ni which are those used in this study.^{7 10 13 18}

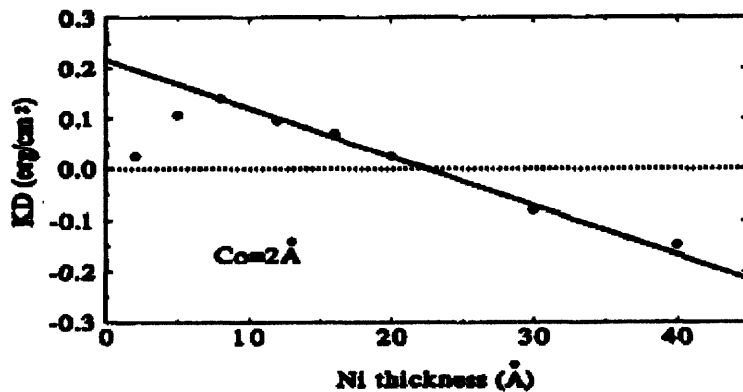


Figure 2: A plot of the total anisotropy constant per unit volume, K , multiplied by the bilayer thickness against the Ni layer thickness of a Co/Ni multilayer 8· (2Å Co/t Å Ni).(Figure from Zhang et al.¹⁸)

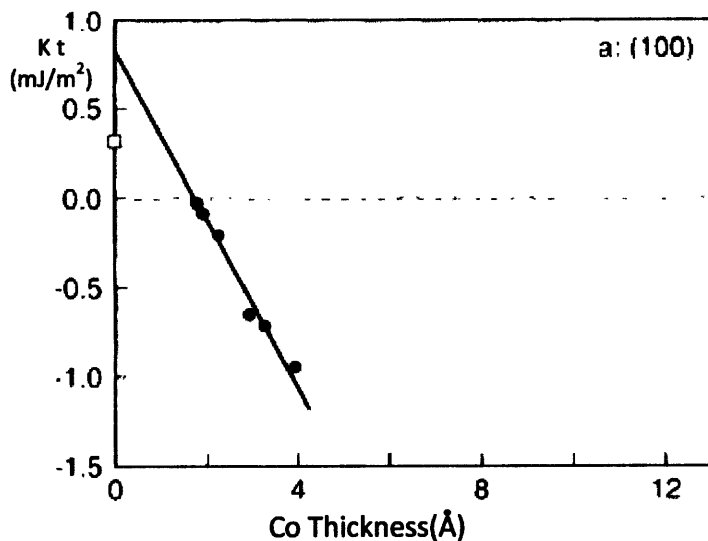


Figure 3: A plot of the total anisotropy constant per unit volume, K , multiplied by the combined thickness of the Ni and Co layers against the Co layer thickness for a Cu(1 0 0)/8Å Ni/Co-Wedge(0-10Å), 1.2 Å/mm/17 Å Ni/7Å Cu/20Å Au) multilayer.(Figure from Johnson et al.¹²)

Orientation

The effect of crystallographic orientation was studied by Johnson et al.¹² The crystallographic orientation of the Co/Ni multilayer is critically important in determining perpendicular magnetic anisotropy.¹² The crystallographic orientation of the multilayer affects the perpendicular anisotropy in three ways.¹² First the crystallographic orientation affects the magnetic interface anisotropy.¹² This is due to different symmetries at the interfaces for different orientations.¹² The crystallographic orientation also affects the magnetic anisotropy by changing the electronic structure of the interface.¹² Lastly the crystallographic orientation affects the magnetic anisotropy by affecting the magnetoelastic anisotropy as different orientations lead to different strains.¹² Daalderop et al. predicted and Johnson et al. showed that the (1 1 1) orientation is optimal for achieving perpendicular magnetic anisotropy in Co/Ni multilayers.^{10 12}

Johnson et al. demonstrated this by growing Ni/Co (wedge)/Ni sandwiches on 100Å copper buffer layers oriented in the (100), (110) and (111) directions and finding the anisotropy with a magneto-optical Kerr effect instrument.¹² The cobalt layer thickness was varied between 0Å and 10Å.¹² It was confirmed that the orientation of the multilayer was the same as the orientation of the copper buffer layers.¹² K was used to quantify the magnetic anisotropy.¹² It was found that only the (1 1 1) orientation had positive values of K .¹² These values of K were also relatively large.¹² Further, for the (1 1 1) orientation K was positive for all the cobalt thicknesses and decreased with increased cobalt thickness but gradually relative to the other orientations and remained positive at the greatest thickness.¹² Both the (1 0 0) and (1 1 0) orientations resulted in 0 values of K or K close to 0 at small cobalt thicknesses.¹² For these orientations K also decreased with thickness becoming more and more negative as expected and described earlier.¹² It is of interest to note that K for the (1 0 0) orientation decreased at a much

faster rate with increased cobalt thickness than K for the (1 1 0) orientation.¹² However, K for both the (1 0 0) and (1 1 0) orientation decreased at a faster rate with increasing cobalt thickness than K for the (1 1 1) orientation.¹²

It was confirmed that both the interface anisotropy and the magnetoelastic anisotropy were affected by the orientation.¹² The interface anisotropy component of K , K_S , was found to be positive and the largest for the (1 1 1) orientation with K_S for the (1 0 0) and (1 1 0) orientations also positive but at roughly half the magnitude of K_S for (1 1 1).¹² The volume anisotropy, K_V , of which the magnetoelastic anisotropy is a component of was found to be highly negative for the (1 0 0) orientation, negative but approximately three times smaller in magnitude for the (1 1 0) orientation and negative but almost six times smaller in magnitude for the (1 1 1) orientation (closer to 0).¹² It was also calculated the magnetoelastic component of the volume anisotropy, K_σ , was positive for the (1 1 1) orientation but negative and of twice the magnitude for the (1 0 0) orientation.¹²

Johnson et al.'s analysis suggests that the (111) orientation is the best choice of orientation for the multilayer's as this orientation is the only one to lead to a positive value of K , and this value of K was also high.¹² This high value of K is a result of the (111) orientation causing interfaces with the highest interface anisotropy as quantified by K_S and is a result of the nature of the strain introduced by (111) orientation causing the highest magnetoelastic anisotropy, as quantified by K_σ .¹² Further, Johnson et al.'s analysis suggests that this can be controlled by the orientation of the buffer layer.¹²

Texture and Buffer Layer

The texture of the Co/Ni multilayers is critical for perpendicular magnetic anisotropy. Texture is the degree to which the crystallographic axes of the oriented grains align.²⁰ Texture most likely affects the perpendicular magnetic anisotropy in the same way as the crystallographic orientation, namely it changes the symmetry at the interfaces and thus affects the interface anisotropy, it changes the electronic structure at the interfaces, and it changes the strain on the multilayer affecting the magnetoelastic anisotropy. The effect of texture has been primarily studied for the (1 1 1) orientation as this orientation as already described, results in strong perpendicular magnetic anisotropy.¹²

It has been found for the (1 1 1) orientation that in order to achieve strong perpendicular magnetic anisotropy the multilayer should be as highly textured as possible. Evidence of this comes from studies where Co/Ni multilayers were deposited onto sputtered buffer layers composed of different materials of various thicknesses which imparted different amount of texture onto the multilayers.^{7 13 18} Broeder et al. performed such a study where he deposited the Co/Ni multilayers on evaporated buffer layers of various materials and thicknesses.¹³(Table 1) (Table 1 from Broeder et al.¹³) While for the buffer layers with very low degrees of (1 1 1) texture there was no correlation between (1 1 1) texture and K , such as for 500Å of Ge, Cr, Ti, Cu, Pd, and 200Å of Au, when there was a large increase in the buffer layer texture, such as for 700Å Au, 200Å Au annealed at 150°C, 200Å Cu on 200Å Au annealed at 150°C and 600Å Au on Si, K increased substantially.¹³(Table 1) Further, for these last four samples as the texture

increased K increased as well.¹³(Table 1) It is important to note that this analysis does not overlap with that in the next paragraph as the three gold buffer layers were processed differently or were deposited on different substrates so there is no trend in these four samples related to the gold buffer layer thickness. However, two of the samples from the study 200Å Au and 700 Å Au are as they were processed the same way. In another study Zhang et al. also showed that Co/Ni multilayers deposited on sputtered silver did not display perpendicular anisotropy while Co/Ni multilayers deposited on sputtered gold did show perpendicular magnetic anisotropy as the gold buffer layer had a greater (1 1 1) texture than the silver buffer layer.⁷

Further evidence for increased texture in multilayers leading to increased K can be found in studies examining the effect of the buffer layer thickness on the K of the Co/Ni multilayer deposited on top.^{13 18} Zhang et al. and Broeder et al., whose study we have already examined, have shown that if Co/Ni multilayers are deposited onto a sputter or evaporated gold buffer layer then as the thickness of the gold buffer layers increases K increases as well.^{13 18} This effect has been attributed to the increasing (1 1 1) texture in the gold buffer layer and thus the Co/Ni multilayers with increasing gold thickness.^{13 18} Broeder et al.'s work show's this trend for two data points as he shows that a 200Å gold buffer layer has a lower degree of texture then a 700Å gold buffer layer and the Co/Ni multilayer deposited on the 200Å gold buffer layer has a much smaller anisotropy constant than the Co/Ni multilayer deposited on the 700Å gold buffer layer.¹³ (Table 1) Zhang et al. meanwhile showed that as the thickness of the gold buffer layer increases K for the multilayers on top transitions from being negative to positive.¹⁸ (Table 2) (Table 2 from Zhang et. al¹⁸) The X-ray diffraction performed on these samples showed that as the thickness of the gold buffer layer increased the (1 1 1) texture of the gold buffer layer increased and also instrumentally X-ray diffraction performed on these samples showed that as the gold buffer layers' degree of (1 1 1) texture increased the degree of (1 1 1) texture of the Co/Ni layers deposited on top increased as well.¹⁸ Hence, the increase in K with increasing gold thickness was attributed to the increased texture of the gold buffer layers which in turn causes increased texture in the Co/Ni multilayers deposited on top.¹⁸

The last piece of evidence for increased texture in the Co/Ni multilayer leading to increased perpendicular anisotropy comes from work done by Naik et al.²¹ Naik et al. etched Si(1 1 1) wafers with either NH₄F or with HF yielding smooth and microscopically rough substrates respectively.²¹ 100Å of Ag, 100Å of Au and the Co/Ni multilayer was subsequently deposited on each buffer layer in that order.²¹ Naik et al. found that the smooth NH₄F etched substrate led to the Ag/Au buffer layer imparting a greater texture on Co/Ni multilayer than the texture imparted on the Co/Ni multilayer by the Ag/Au buffer layer deposited on the rough HCl etched substrate.²¹ In turn the multilayer deposited on the buffer layer on the NH₄ etched substrate had a greater K than the multilayer deposited on the buffer layer deposited on the HCl etched substrate.²¹

It is of interest to note that a group recently found perpendicular anisotropy in Co/Ni multilayers deposited on a silicon substrate with no metal buffer layer.⁴ The group did not explain their finding however or describe the orientation of the silicon substrate.⁴

Thus it is clear that to maximize perpendicular anisotropy Co/Ni multilayers should have as high a (1 1 1) texture as possible. This can be achieved by properly selecting the (1 1 1) buffer layer material, the buffer layer thickness and the roughness of the substrate the buffer layer is deposited on.

Table 1: The total magnetic anisotropy per unit volume, K , of Co/Ni multilayers deposited on buffer layers of varying materials and thicknesses. The degree of (1 1 1) texture of each buffer layer as determined by the intensity of the buffer layer (1 1 1) peak from X-ray diffraction measurements performed on each sample is also shown. Table from Broeder et al. with two samples made under different deposition conditions edited out as they are not comparable.¹³

Substrate	Underlayer	$K(\text{kJ/m}^3)$	$H_C(\text{kA/m})$	/111
Glass	None	-400	-	2
Glass	500 Å Ge	-80	-	2
Glass	500 Å Cr	-80	-	7
Glass	500 Å Ti	80	45	6
Glass	500 Å Cu	85	34	4
Glass	500 Å Pd	100	35	3
Glass	200 Å Au	150	40	8
Glass	700 Å Au	330	67	20
Glass	200 Å Au (Annealed at 150°C)	420	70	49
Glass	200 Å Au/200Å Cu (Annealed at 150°C)	430	130	69
Si	600 Å Au	680	40	100

Table 2: The effect of the Au buffer layer thickness on the total magnetic anisotropy per unit volume, K , of the Co/Ni multilayers deposited on top. (Table from Zhang et al.¹⁸)

Sample No.	Au Buffer Layer Thickness (Å)	$K(\text{kJ/m}^3)$
Au-1	50	-37
Au-2	150	-3
Au-3	250	50
Au-4	350	60
Au-5	450	78
Au-6	550	89
Au-7	650	100

Effect of Deposition Rate

Only one group has examined the effect of deposition rate. For Co/Ni multilayers evaporated at room temperature on evaporated gold buffer layers K increased slightly less than linearly with increased deposition rate from $K \approx 190 \text{ kJ/m}^3$ at a Co deposition rate of $.1 \text{ \AA/s}$ and an Ni deposition rate of $.2 \text{ \AA/s}$ to $K \approx 405 \text{ kJ/m}^3$ at a Co deposition rate of 1 \AA/s and an Ni deposition rate of 2 \AA/s .¹³ Further, when this experiment was repeated at $-196 \text{ }^\circ\text{C}$, K remained fairly constant with the deposition rate at approximately $K \approx 450 \text{ kJ/m}^3$.¹³ Two explanations were proposed for this.¹³ The first is that at faster deposition rates the Co/Ni interfaces are better defined which leads to higher interface anisotropy and a more ideal interface electronic structure.¹³ The second explanation is that a higher deposition rate leads to a greater strain in the deposited film which results in magneto-elastic anisotropy which increases K .¹³ However, no characterization was performed on these films and the method of increasing the deposition rate was not described so it is difficult to say if their explanation was valid.¹³

There were no other studies of how K varied with deposition rate. It is difficult to construct deposition rate- K curves for other deposition methods by assembling data from different papers as the buffer layers, layer thicknesses and numbers of layers are different in all the papers. Further even if we neglect these differences few studies list their deposition rates and the details of their deposition and we would not have enough data points for each method. However we will still summarize the deposition methods and rates in studies of Co/Ni multilayers with perpendicular magnetic anisotropy. (Table 3)

Table 3: Deposition methods and deposition rates used to deposit Co/Ni multilayers with perpendicular magnetic anisotropy.

Deposition Method	Deposition Rate	Study
Evaporation	Co: $.1 \text{ \AA/s}$ - 1 \AA/s Ni: $.2 \text{ \AA/s}$ - 2 \AA/s (Highest deposition rate for Co and Ni yields highest value of K)	Broeder et al. ^{13,19} (2) and Daalderop et al. ¹⁰
Molecular Beam Epitaxy	Co: $.5 \text{ \AA/s}$ Ni: $.5 \text{ \AA/s}$	Naik et al. ²¹
Molecular Beam Epitaxy	Co: $.1 \text{ \AA/s}$ Ni: $.1 \text{ \AA/s}$	Bloemen et al. ²²
Magnetron Sputtering	Co: $.9 \text{ \AA/s}$ Ni: 1.3 \AA/s	Zhang et al. ¹⁸

Roughness and Interface Diffusion

While there is some overlap with this section and other sections it is important enough to consider separately. It has been suggested that Co/Ni multilayers have a lower K_s , the interface anisotropy, when the interfaces are rough.¹⁷ This would most likely be the case as roughness

would change the symmetry at the Co/Ni interfaces and the electronic structure at the interfaces which are both causes of the magnetic interface anisotropy. Interface diffusion could also lower K_S .^{13 17 18} This could occur as interface diffusion results in alloying the interface which again changes the symmetry, electronic structure, and the general structure of the interface.

Effect of Deposition Method

The deposition method appears to primarily affect K_S the interface anisotropy. It has been observed that both MBE and electron beam evaporated Co/Ni multilayers yield a higher K_S than evaporated Co/Ni multilayers.^{17 18} It has been suggested that K_S is larger for e-beam evaporated multilayers than for sputtered multilayers as evaporated multilayers have better defined interfaces.¹⁸ Evaporated multilayers could have better defined interfaces than sputtered multilayers as evaporated atoms have a much lower kinetic energies than sputtered atoms.²³ For example, a typical energy for a sputtered atom is 10eV and a typical energy of an evaporated atom is less than .2eV.²³ Hence, sputtered atoms deposited onto a surface have much greater energy and are more likely to diffuse at interfaces forming alloys. This leads to less defined interfaces for sputtered multilayers and less defined interfaces due to alloy formation at the interfaces could decrease K_S as already described. A similar argument could apply to MBE as it is in essence a slower form of evaporation.²⁴ In addition it has been suggested that MBE yields multilayers with a higher K_S than sputtering as MBE deposited multilayers and thus their interface are not as rough as interfaces in sputtered multilayers.¹⁷

Effect of Substrate Temperature During Deposition

For evaporated Co/Ni multilayers it has been found that as the substrate temperature during deposition increases from 20°C to 800°C K_S decreases exponentially from 300 kJ/m³ to 45 kJ/m³.¹³ Two causes have been proposed for this. The first cause is proposed to be that at high temperatures there is diffusion at the interfaces reducing the magnetic interface anisotropy and changing the electronic structure of the interfaces.¹³ The second cause is proposed to be reduced strain in multilayers for deposition at higher temperatures resulting in less magneto-elastic anisotropy.¹³ Further explanations for these proposals is that at higher temperatures the atoms deposited at the interface have increased energy and will diffuse leading to poorly defined interfaces and alloying which affects the electronic structure at the interface and changes the symmetry at the interface, which as we have already discussed are critical for perpendicular magnetic anisotropy.¹³ Also, the additional energy imparted on the deposited atoms by the heated substrate could allow them to rearrange to reduce strain, hence reducing the magnetoelastic anisotropy.¹³ Again it is important to note that no characterization was done so both explanations for the trend observed are tentative.¹³

Chapter 3: Experimental

Co/Ni Multilayer Preparation

The multilayers examined were prepared in an ultrahigh vacuum system with the pressure prior to deposition at less than $6 \cdot 10^{-8}$ torr. The samples were prepared on pieces of Si wafers oxidized at atmospheric conditions. The substrates were cleaned in acetone and rinsed with isopropyl alcohol prior to deposition. Also immediately prior to deposition the substrates were blown clean with nitrogen gas. The pieces of wafer used were typically square with an edge length of .5 cm.

Two methods were used for the deposition of the multilayers, Ion Beam Sputtering and DC magnetron sputtering. DC magnetron sputtering was performed with Argon gas at 1 mtorr. The target bias voltage was .1kW, the plasma current was approximately .61A for the tantalum and .46A for the gold, and the plasma voltage was 172V for the tantalum and 239V for the gold. Ion beam sputtering was performed with Argon gas at $4 \cdot 10^{-5}$ torr with a cathode current of 6.8A, a cathode voltage of 7.9V, a discharge current of 1.0A, a discharge voltage of 39.9V, a beam current of 35.5 mA, a beam voltage of 999V, an accelerator current of 1.1 mA and an accelerator voltage of 200V. DC magnetron sputtering was used to deposit tantalum and gold and ion beam sputtering was used to deposit copper, cobalt and nickel. For DC magnetron sputtering each element had a separate sputtering gun. For ion beam sputtering there was one gun which contained four targets that could be rotated between. The deposition rate for each element for both the DC magnetron sputtering and the ion beam sputtering was determined by calibration. The calibration process consisted of marking a 6" wafer with a grid of lines with a permanent marker and depositing a thick film (approximately 40nm for DC magnetron sputtering and 20nm for ion beam sputtering) of one of the materials on the wafer. The permanent marker and the deposited material on top were subsequently lifted off by immersion in acetone for 15 minutes followed by sonication and then rinsed in isopropyl alcohol. The height of the resulting steps was then recorded with atomic force microscopy or a Tencor P-16 Surface Profilometer at various points at the center of the wafer. The average thickness was divided by the deposition time to determine the deposition rate. The deposition rates of the elements sputtered can be found in the tables below. (Table 4 and Table 5)

Table 4: The deposition rate of elements deposited by ion beam sputtering.

Element	Deposition Rate ($\text{\AA}/\text{s}$)
Co	.260
Ni	.253
Cu	.380

Table 5: The deposition rate of elements deposited by direct current magnetron sputtering.

Element	Deposition Rate (Å/s)
Ta	1.97
Au	8.97

The multilayers that were prepared can be found in Table 6. (Table 6) Each multilayer consisted of sequentially, an adhesion layer, ten 2Å Cobalt 8Å Nickel or 2Å Cobalt 4Å Nickel bilayers and a 20Å Au capping layer to prevent oxidation. The Ta adhesion layers were deposited by sputtering as were the gold capping layers. The cobalt/nickel bilayers were deposited by ion beam sputtering. The buffer layer was deposited by either ion beam sputtering or DC magnetron sputtering as the copper buffer layers was deposited with ion beam deposition and the gold buffer layers were deposited with DC magnetron sputtering. Prior to the deposition of the adhesion layer, the buffer layer, the first layers of cobalt and nickel, and the capping layer a 2:00 minute pre-sputter was performed. Prior to the deposition of each of the layers in the other 9 cobalt/nickel bilayers a 15s pre-sputter was performed.

Table 6: Specifications of the thin film Co/Ni multilayers fabricated to try to achieve perpendicular magnetic anisotropy. The Co/Ni layers were deposited by ion beam sputtering and the buffer layers were deposited by ion beam sputtering or direct current (DC) magnetron sputtering.

Sample Number	Adhesion Layer	Buffer Layer	Co/Ni Layers	Capping Layer	Buffer Layer Deposition Method
a	20Å Ta	30Å Au	10x(Co 2Å/Ni 4Å)	20Å Au	DC Magnetron Sputtering
b	20Å Ta	30Å Au	10x(Co 2Å/Ni 8Å)	20Å Au	DC Magnetron Sputtering
c	20Å Ta	650Å Au	10x(Co 2Å/Ni 8Å)	20Å Au	DC Magnetron Sputtering
d	20Å Ta	200Å Cu	10x(Co 2Å/Ni 8Å)	20Å Au	Ion Beam Sputtering
e	20Å Ta	400Å Cu	10x(Co 2Å/Ni 8Å)	20Å Au	Ion Beam Sputtering

X-Ray Diffraction Characterization

X-Ray diffraction was performed on all the samples on a Rigaku RU300 diffractometer with a Cu Ka radiation source at 50 mA and 300V. The scans were performed between 20° and 80°, at 2.0° per minute, with a step size of .02, and a ψ of 1°.

Atomic Force Microscopy Characterization

Atomic force microscopy was performed in tapping mode on a Veeco Metrology Nanoscope IV Scanned Probe Microscope on 1 μm by 1 μm areas. Roughness was analyzed with the Nanoscope Software Version 5.30r3sr3 by first performing first order flattening and then calculating the root mean square roughness (rms).

Magnetic Characterization

Magnetic characterization was performed with two instruments an alternating gradient field magnetometer (AGFM) and a vibrating sample magnetometer. The AGM was made by Princeton Applied Research and was calibrated with a 511 μemu piece of nickel foil. Magnetic hysteresis loops were measured for each sample with the sample surface both parallel to the applied field (in plane) and the sample surface perpendicular to the applied field (out of plane) between applied fields of 3 kOe and -3 kOe in 20 Oe increments.

Vibrating sample magnetometry was performed on each of the samples with a ADE Magnetics vibrating sample magnetometer. Magnetic hysteresis loops were measured for each sample with the sample surface both parallel to the applied field (in plane) and the sample surface perpendicular to the applied field (out of plane) as for the AGM. Hysteresis loops were measured between applied fields of -10 kOe and 10 kOe.

Chapter 4: Results and Discussion

The X-Ray diffraction data on the samples showed that each samples buffer layer had (1 1 1) texture however the extent of the (1 1 1) texture strongly depended on the buffer layer material and the buffer layer thickness.(Figure 4) Both of the samples prepared with the 30Å gold buffer layer, 20Å Ta/30Å Au/10x(Co 2Å/Ni 4Å)/ 20Å Au and 20Å Ta/30Å Au/10x(Co 2Å/Ni 8Å)/ 20Å Au, samples a and b respectively, had broad (1 1 1) gold peaks of fairly low intensity in their X-ray diffraction patterns. This suggests that the deposited gold buffer layer had a low degree of (1 1 1) texture and thus the multilayer deposited on top also had low degree of (1 1 1) texture. The reason for the low degree of (1 1 1) texture observed is most likely the thinness of the gold layer. This was confirmed to some extent by the sample with the 650Å gold buffer layer, sample c. The sample with the 650Å gold buffer layer 20Å Ta/650Å Au/10x (Co 2Å/Ni 8Å)/ 20Å Au had a very strong and sharp (1 1 1) gold peak in its diffraction pattern suggesting a high degree of (1 1 1) texture in the buffer layer and thus the multilayer. As samples b and c were exactly the same except for the difference in the thickness of the gold buffer layers the increase in the degree of (1 1 1) texture for sample c relative to sample b can be attributed to the increased thickness of the gold buffer layer. From, this we can draw the conclusion that increased gold thickness improves the degree of (1 1 1) texture of the gold and thus the (1 1 1) texture of the multilayer deposited on top. This confirms the results of Zhang et al. which we described earlier which showed that as the thickness of the gold buffer layer increases, the degree of (1 1 1) texture of the gold buffer layer increases and the degree of (1 1 1) texture of the multilayer increases with the increase in the degree of (1 1 1) texture of the gold buffer layer.¹⁸

The X-ray diffraction pattern for the sample with the 200Å copper buffer layer, 20Å Ta/200Å Cu/10x(Co 2Å/Ni 8Å)/ 20Å Au, sample, d, showed a broad very low intensity (1 1 1) copper peak suggesting that the 200 Å Cu buffer layer had a very low degree of (1 1 1) texture and thus that the multilayer deposited on top had a very low degree of (1 1 1) texture. The X-ray diffraction pattern for the sample with the 400Å copper buffer layer, the 20Å Ta/400Å Cu/10x(Co 2Å/Ni 8Å)/ 20Å Au sample, sample e, had a moderately sharp and low/moderately intense (1 1 1) copper peak suggesting that it had a low to moderate degree of (1 1 1) texture and thus that the multilayer deposited on top had a low to moderate degree of (1 1 1) texture. The increased peak intensity and sharpness for sample e compared to sample d can be attributed to the increased thickness of the Cu buffer layer in sample e as the samples are identical otherwise. Again as for gold this suggests that a thicker copper buffer layer has a higher (1 1 1) texture and thus the multilayer on top also has a higher degree of (1 1 1) texture.

The X-ray diffraction patterns also allow us to an extent to compare the degree of (1 1 1) texture of the gold and copper buffer layers. The 30Å gold buffer layer samples had more intense (1 1 1) peaks than the 200Å copper buffer layer sample despite the greater thickness of the

copper buffer layer. This suggests that gold has a better degree of (1 1 1) texture for a given thickness. Also, the 650Å gold buffer layer sample has a much more intense and sharp (1 1 1) peak than the 400Å Copper buffer layer sample. While the increased intensity and sharpness could be from the greater thickness of the gold buffer layer, the intensity is so much greater for the sample with the gold buffer layer that most likely the greater intensity of the peak is coming in part from the higher degree of (1 1 1) texture in gold. This is also confirmed by Broeder et al. who found that a 200Å gold buffer layer with a Co/Ni multilayer deposited on top had a greater degree of (1 1 1) texture than a 500Å copper buffer layer with a Co/Ni multilayer deposited on top.¹³(Table 1) It is important to note that the gold and copper buffer layers were prepared by two different methods at two very different deposition rates. The gold buffer layers were prepared by DC magnetron sputtering at a relatively high deposition rate and the copper buffer layers were prepared by IBS at a relatively low deposition rate. Hence the differences in the degrees of texture for the copper and gold films could also be the result of the different deposition methods.

To summarize our X-ray diffraction data we have shown that samples a and b have buffer layers with a low degree of (1 1 1) texture due to thinness of the gold film and sample d has a buffer layer with a low degree of (1 1 1) texture due to thinness of the buffer layer and the relatively low degree of (1 1 1) texture for IBS copper. Sample e has a buffer layer with a low to moderate degree of (1 1 1) texture as the copper buffer layer is fairly thick and sample c has a buffer layer with a high degree of (1 1 1) texture due to the thickness of the gold buffer layer and the relatively high degree of (1 1 1) texture for sputtered gold. The general trends determined from the X-ray diffraction data is that an increased copper or gold buffer layer thickness leads to an increased intensity and sharpness in the (1 1 1) peak and thus a greater degree of (1 1 1) texture. Also, DC magnetron sputtered gold has a greater degree of (1 1 1) texture than IBS sputtered copper. This data is useful as it suggests the level of (1 1 1) texture in the Co/Ni multilayers deposited on top of the different buffer layers which has an important effect on K and will be used to interpret the magnetic data.

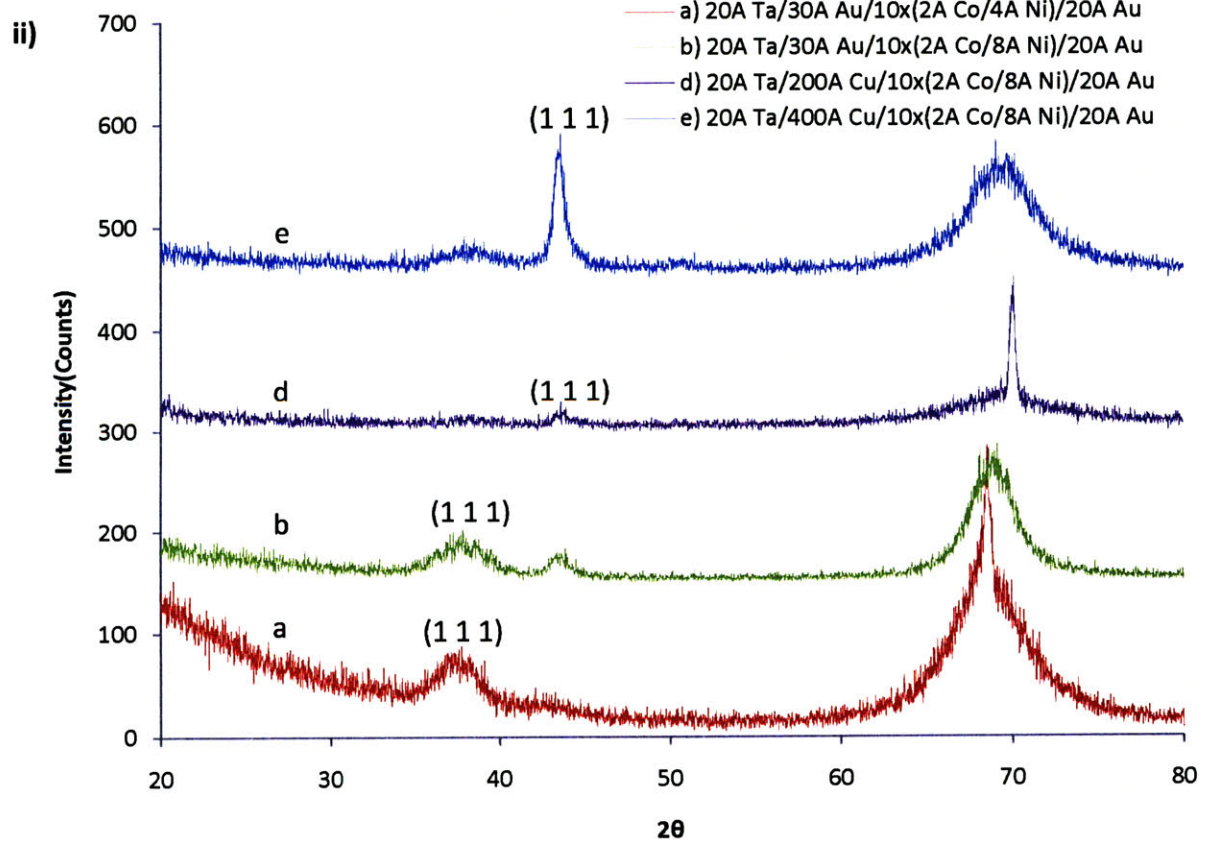
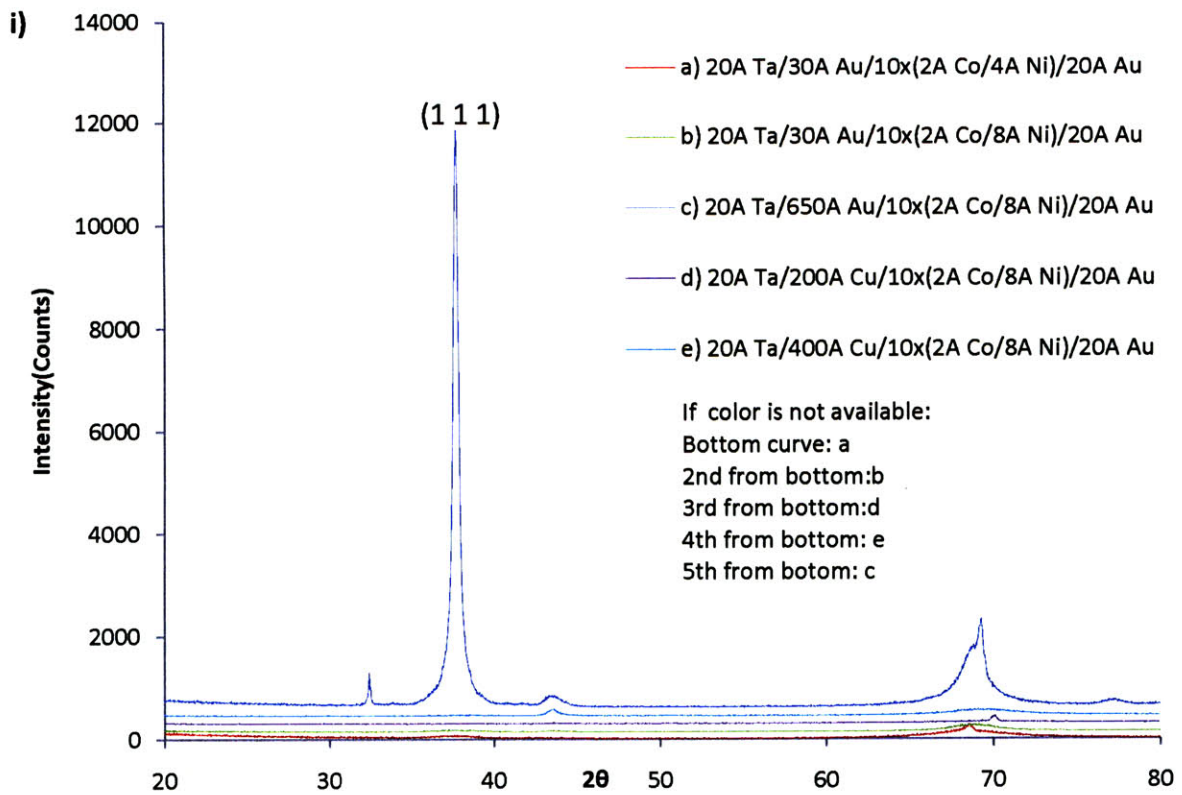


Figure 4: X-ray diffraction patterns of Co/Ni multilayers deposited on different buffer layers. The composition of each sample and the buffer layer it is deposited on is listed in the legend. i) shows the X-ray diffraction patterns of all the samples on the same graph with the patterns shifted vertically to allow for easy viewing. ii) shows the X-ray diffraction patterns of all the samples on the same graph except for sample c with the patterns shifted vertically to allow for easy viewing. Sample c) was removed in ii) as the magnitude of its (1 1 1) peak was so high that plotting it on the same axes with the other four patterns made the other four patterns difficult to examine effectively. The (1 1 1) peak of the buffer layer of each sample is labeled.

The atomic force microscopy on the surface of each sample allowed us to characterize the roughness of each sample by examining the microscopy images and by analyzing the roughness mathematically with the Nanoscope software via the root-mean-square (RMS) roughness parameter. (Figure 5 and Table 7) A visual examination of the microscopy images of samples a and b shows a similar level of roughness on a fine length scale although the roughness in image b seems to be on a smaller length scale than the roughness in image a. The RMS roughness was calculated to be .179 nm for sample a and .214 nm for sample b both very low values. Hence samples a and b have very little roughness and the roughness is on a fine length scale. The results show that the roughness of Co/Ni multilayers deposited on a very thin 30Å gold layer is very low and on a small length scale. This in turn suggests that both the thin 30Å gold buffer layer is fairly smooth and that the deposited Co/Ni multilayer is fairly smooth and would be so if deposited on a flat surface separately. This is important to note as if a different buffer layer with a Co/Ni multilayer on top yields a rough surface the roughness can be attributed to the buffer layer and not the Co/Ni multilayer. The increased roughness of sample b compared to sample a can be attributed to the fact that sample b has 8Å nickel layers while sample a has 4Å nickel layers. Thus, the increased thickness of sample b could be the cause of its slightly increased roughness. This also suggests that the roughness of multilayer does contribute to total roughness and not only is the buffer layer responsible for the roughness. This also suggests that increasing the thickness of the layers in the multilayer will generally increase the roughness.

Sample c was very rough as determined from the microscopy images and the RMS roughness. From the microscopy images the roughness was present on a relatively broad length scale as well as on a fine length scale. From the microscopy images the roughness was also much greater for sample c than for samples a and b which is especially evident if one notices the depth scale bar is on 3.7 times larger scale for sample c than for samples a and b. The RMS roughness was .749 nm which confirms that the sample is very rough and that it is much rougher than samples a and b. The explanation for the increased roughness is the thickness of the gold buffer layer as this is the only difference between samples b and c. The significantly thicker gold buffer layer must be causing the increased roughness of sample c. Thus, this suggests that a thick gold buffer layer on the order of 650Å is very rough and in general the greater thickness the sputtered gold buffer layer the greater the roughness of the buffer layer and the Co/Ni multilayer on top.

The microscopy images of samples d and e reveal low surface roughness with roughness on a small length scale. The roughness appears to be slightly greater than samples a and b but the length scale of the roughness is smaller. The roughness of sample e appears to be slightly larger

than the roughness of sample d and the length scale of the roughness is slightly greater. The computed RMS roughness's were .235 nm for sample d and .289 nm for sample e. Thus, samples d and e have very little roughness. Further, the RMS values confirm that samples d and e are rougher than samples a and b but much less rough than sample c. The fact that sample e is much less rough than sample c suggests that IBS copper provides a smoother buffer layer than DC magnetron sputtered gold as while the thickness of the copper in sample e is much smaller than the thickness of the gold in sample c, 400Å of copper is thick enough such that if copper was as rough as gold its roughness would be much closer to the roughness of gold at 600Å than the data shows. This is further supported by the fact that the RMS roughness of copper only increases by .054 nm when the copper buffer layer thickness is increased from 200Å to 400Å between samples d and e. Thus an increase in the copper buffer layer to 650Å is unlikely to raise its RMS roughness anywhere near that of a sample of with a 650Å gold buffer layer with a RMS roughness of .749. Thus, IBS copper is far less rough than DC magnetron sputtered gold. The slight increase in roughness between samples d and e due to the thicker copper buffer layer does show that the roughness of the copper buffer layer and the Co/Ni multilayer on top does increase when the copper buffer layer is increased in thickness.

In conclusion our AFM data shows that the multilayers deposited on thin 30Å gold buffer layers, a and b, are very smooth, the multilayer deposited on a thicker 650Å gold buffer layer is very rough, and the multilayers deposited on copper buffer layers are smooth but not as smooth as the multilayers deposited on the 30Å gold buffer layers. Our data also shows that at greater thicknesses DC magnetron sputtered gold is much rougher than IBS copper and that the roughness of a buffer layer and hence the multilayer on top generally increases with buffer layer thickness. Lastly, our data shows that when thickness of a layer in the multilayer increases the roughness of the multilayer increases slightly as well.

Table 7: The roughness of thin film Co/Ni multilayers deposited on various buffer layers as measured by AFM. The Co/Ni layers were deposited by ion beam sputtering (IBS) and the buffer layers were deposited by IBS or direct current (DC) magnetron sputtering as is indicated for each sample in the table. The 20 Å Ta adhesion layer and the 20Å Au capping layer for each sample are not included in the sample descriptions.

Sample Label	Sample Structure	Method of Buffer Layers Deposition	RMS Roughness(nm)
a	30Å Au/10x(Co 2Å/Ni 4Å)	DC Magnetron Sputtering	.179
b	30Å Au/10x(Co 2Å/Ni 8Å)	DC Magnetron Sputtering	.214
c	650Å Au/10x(Co 2Å/Ni 8Å)	DC Magnetron Sputtering	.749
d	200Å Cu/10x(Co 2Å/Ni 8Å)	IBS	.235
e	400Å Cu/10x(Co 2Å/Ni 8Å)	IBS	.289

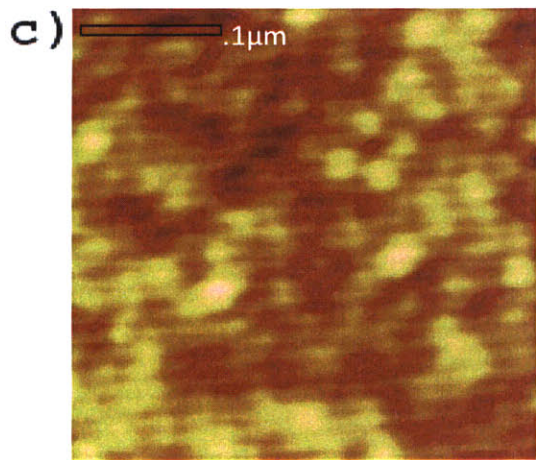
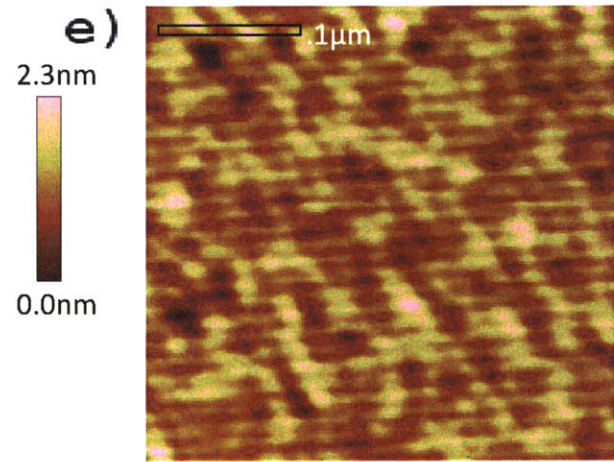
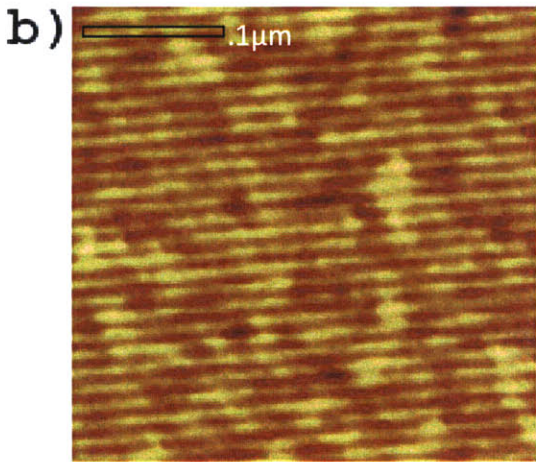
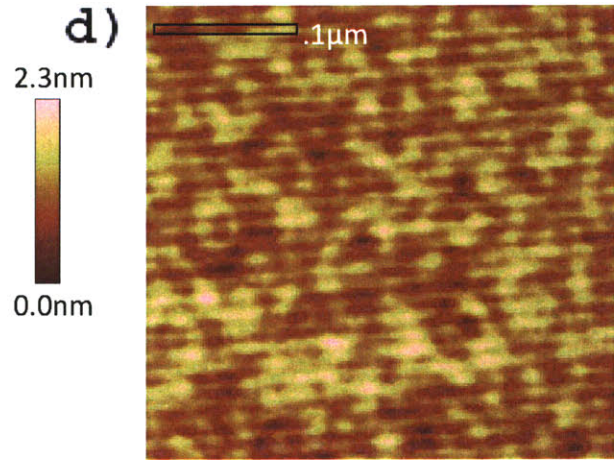
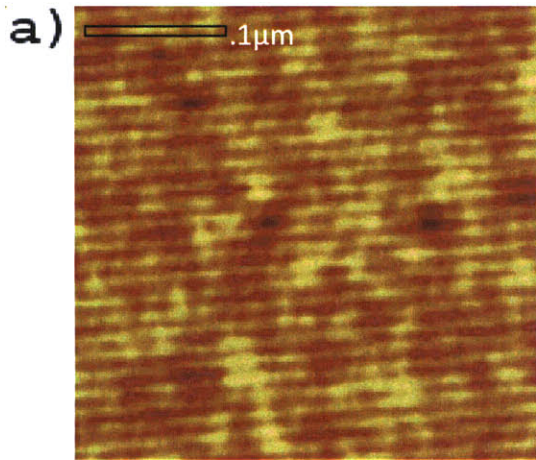


Figure 5: Atomic force microscopy (AFM) images of Co/Ni multilayers deposited on various buffer layers. Each depth scale bar applies to the images in that row. The structure of each sample analyzed is listed below.

a) 20Å Ta/30Å Au/10x(2Å Co/4Å Ni)/20Å Au
 b) 20Å Ta/30Å Au/10x(2Å Co/8Å Ni)/20Å Au
 c) 20Å Ta/650Å Au/10x(2Å Co/8Å Ni)/20Å Au
 d) 20Å Ta/200Å Cu 10x(2Å Co/8Å Ni)/20Å Au
 e) 20Å Ta/400Å Cu/10x(2Å Co/8Å Ni)/20Å Au

8.5nm
0.0nm

Magnetic data was taken in the form of hysteresis loops on the sample by AGM and VSM. Unfortunately due to instrument problems the magnitudes of the magnetic moments measured were not valid however the shapes of the hysteresis loops were which we confirmed by performing both AGM and VSM. The AGM and VSM data showed that all the samples we deposited clearly had in-plane magnetic anisotropy. (Figure 6 and Figure 7) Hence the rest of our discussion of the magnetic data will focus on understanding why the samples had in plane magnetic anisotropy instead of out of plane magnetic anisotropy using the XRD and AFM characterization data we obtained and our literature review. At first we will specifically analyze possible reasons for why each sample has in plane magnetic anisotropy instead of out of plane magnetic anisotropy and then we will analyze possible general reasons for why we obtained in plane anisotropy that would affect all the samples.

Samples a and b most likely had parallel magnetic anisotropy instead of perpendicular magnetic anisotropy as the 30Å buffer layer for these samples was too thin and had a small degree of (1 1 1) texture. The small degree of (1 1 1) texture in the 30Å gold buffer layers was seen in the XRD patterns. Thus, the gold buffer layer imparted only a small degree of (1 1 1) texture on Co/Ni multilayer deposited on top and thus its interfaces. As the (1 1 1) texture at the interfaces of Co/Ni multilayers is critical for perpendicular magnetic anisotropy as we have already discussed in the literature review, the lack of (1 1 1) texture at the interfaces results in parallel magnetic anisotropy.^{7 13 18} The low degree of (1 1 1) texture was probably the cause and not the roughness as the samples were fairly smooth as shown by AFM. Our result was confirmed to some extent by Zhang et al. who, as we have already discussed to some extent, found that for Co/Ni multilayers sputtered onto gold buffer layers on glass as the thickness of the gold buffer layers increased K increased as well transitioning from negative to positive at a gold buffer layer thickness between 150Å and 250Å.¹⁸ Zhang et al. also found with XRD that as the thickness of the gold buffer layer increased the (1 1 1) texture of the gold buffer layer and the (1 1 1) texture of the Co/Ni multilayers layer increased as well and thus proposed that multilayers deposited onto gold buffer layers of less than 150Å have parallel magnetic anisotropy as thin gold buffer layer have a low degree of (1 1 1) texture and hence the Co/Ni multilayers on top also have a low degree of (1 1 1) texture which is not sufficient for perpendicular magnetic anisotropy.¹⁸

Sample c most likely had parallel magnetic anisotropy as it was too rough. The 650Å gold buffer layer of sample c caused it to have a RMS roughness at the surface of .749nm which is equivalently 7.49Å. Compared to the thickness of the Co and Ni layers, 2Å and 8Å respectively, this is a very high level of roughness. This surface roughness indicates that the interfaces of the Co/Ni multilayers were very rough as well affecting the symmetry at the interfaces and hence the magnetic interface anisotropy, the electronic structure at the interfaces and the magnetoelastic anisotropy. All of these are the principal causes of perpendicular magnetic anisotropy and hence their disruption by roughness leads to in plane anisotropy as was observed in sample c. The roughness was probably the cause of the parallel magnetic anisotropy and not the degree of (1 1 1) texture as the degree (1 1 1) texture for sample c was good as shown

by XRD. Roughness was cited as a cause of decreased perpendicular anisotropy by Johnson et al.¹⁷

Sample d most likely had parallel magnetic anisotropy as the 200Å copper buffer layer had a very low degree of (1 1 1) texture even lower than samples a and b. Thus, in plane anisotropy would occur for the same reasons as for samples a and b. The sample was fairly smooth as shown by AFM so roughness probably did not cause the parallel magnetic anisotropy.

The parallel magnetic anisotropy in sample e is more difficult to explain with the AFM and XRD data. Sample e had a low to moderate degree of (1 1 1) texture and low RMS roughness. It is possible that the degree of (1 1 1) texture was too low and the RMS roughness was still too high in the Co/Ni multilayer, however, it is difficult to speculate in this case if either of these factors is responsible for the observed magnetic behavior as we did not determine what degree of (1 1 1) texture and smoothness is required for perpendicular magnetic anisotropy. It is of interest to note that Broeder et al. found that a Co/Ni multilayer deposited onto a 500Å copper buffer layer on glass had a low degree of (1 1 1) texture and a positive but low value of K suggesting that maybe for our 400Å Cu buffer layer sample the low to moderate degree of (1 1 1) texture causes K to be fairly low and when this is combined with another factor such as interface diffusion it makes K negative.¹³ Furthermore Broeder et al.'s work definitely suggests that the low degree of (1 1 1) texture of a 200Å copper buffer layer will result in a Co/Ni multilayer deposited on top with a too low degree of (1 1 1) texture for perpendicular magnetic anisotropy.¹³

Now we will consider a general reason for why perpendicular magnetic anisotropy was not present in the multilayers. The multilayers were deposited by sputtering. Deposited atoms in sputtering have high energies and upon being deposited on a surface can diffuse.²³ Thus, when a multilayer is deposited the deposited atoms due to their high energy can diffuse at the interface forming alloys. The diffusion and alloy formation at the interfaces of Co/Ni multilayers disturbs the symmetry of the interfaces and hence the magnetic interface anisotropy, the electronic structure at the interfaces, and the magnetoelastic strain, all causes of perpendicular magnetic anisotropy. Thus the diffusion at the interfaces in sputtering can lead to parallel magnetic anisotropy instead of perpendicular magnetic anisotropy.

Evidence to support this is that as discussed in the literature review MBE deposited Co/Ni multilayers and evaporated Co/Ni multilayers have higher K_S values than sputtered Co/Ni multilayers.^{17 18} MBE and evaporation deposited multilayers have better defined interfaces as MBE and evaporated atoms have much less energy, as described in the literature review, than sputtered atoms and hence the atoms have less energy at the surface on deposition and diffuse to a lesser extent at the interfaces leading to less alloying at the interfaces.²³ Further, MBE and evaporation deposited multilayers have been proposed to have a higher K_S due to less roughness.¹⁷ Thus, the higher K_S for MBE and evaporated Co/Ni multilayers as compared to sputtered multilayers can be attributed to less diffusion at the interfaces and decreased roughness.

What may have occurred in our case is that as just discussed we may have had too much diffusion and alloy formation at the interfaces as our multilayers were deposited by sputtering

and hence this diffusion and alloy formation resulted in the parallel magnetic anisotropy. Alternatively it is possible that diffusion and alloy formation decrease the perpendicular magnetic anisotropy and when this is combined with issues with the buffer layer such as roughness and a low degrees of (1 1 1) texture this led to parallel magnetic anisotropy.

It is of interest to note that as already mentioned perpendicular magnetic anisotropy has been achieved in Co/Ni multilayers prepared with DC magnetron sputtering by Zhang et al.^{7 18} Possible reasons for their obtaining perpendicular magnetic anisotropy with sputtering and us obtaining parallel magnetic anisotropy could be that we used IBS and they used DC magnetron sputtering under different conditions which most likely produces multilayers with different properties.^{7 18} Also, our deposition rate for Co was approximately 3.6 times slower than theirs and our deposition rate for Ni was 5.2 times slower than theirs and it is difficult to say what effect that could have had on the multilayers.^{7 18} In addition the difference may have been in the roughness of their buffer and multilayers and their degree of (1 1 1) texture.

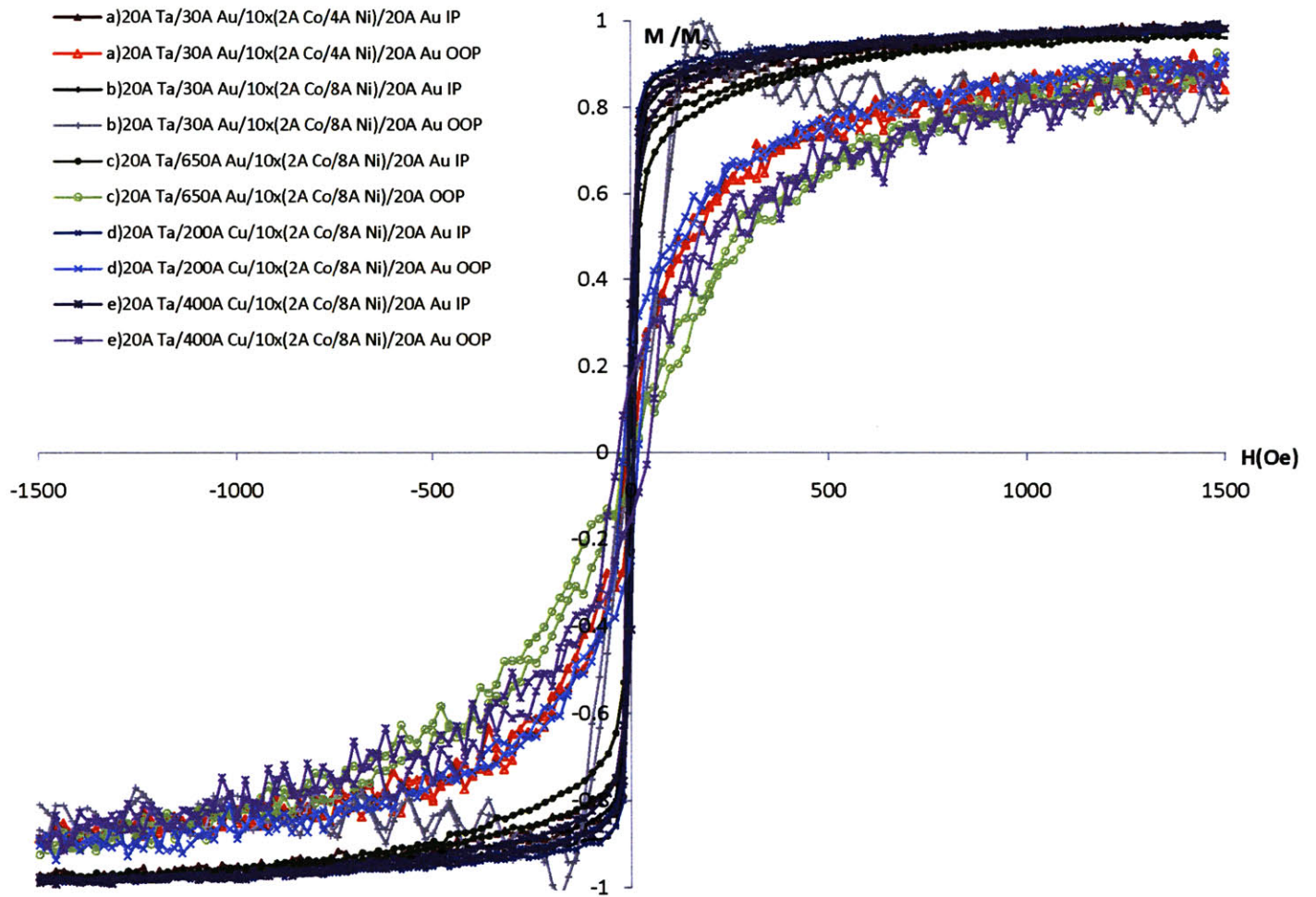


Figure 6: In plane and out of plane magnetic hysteresis loops for various Co/Ni multilayers on gold or copper buffer layers as measured by alternating gradient magnetometry (AGM). The sample compositions are labeled. IP stand for in plane and OOP stand for out of plane.

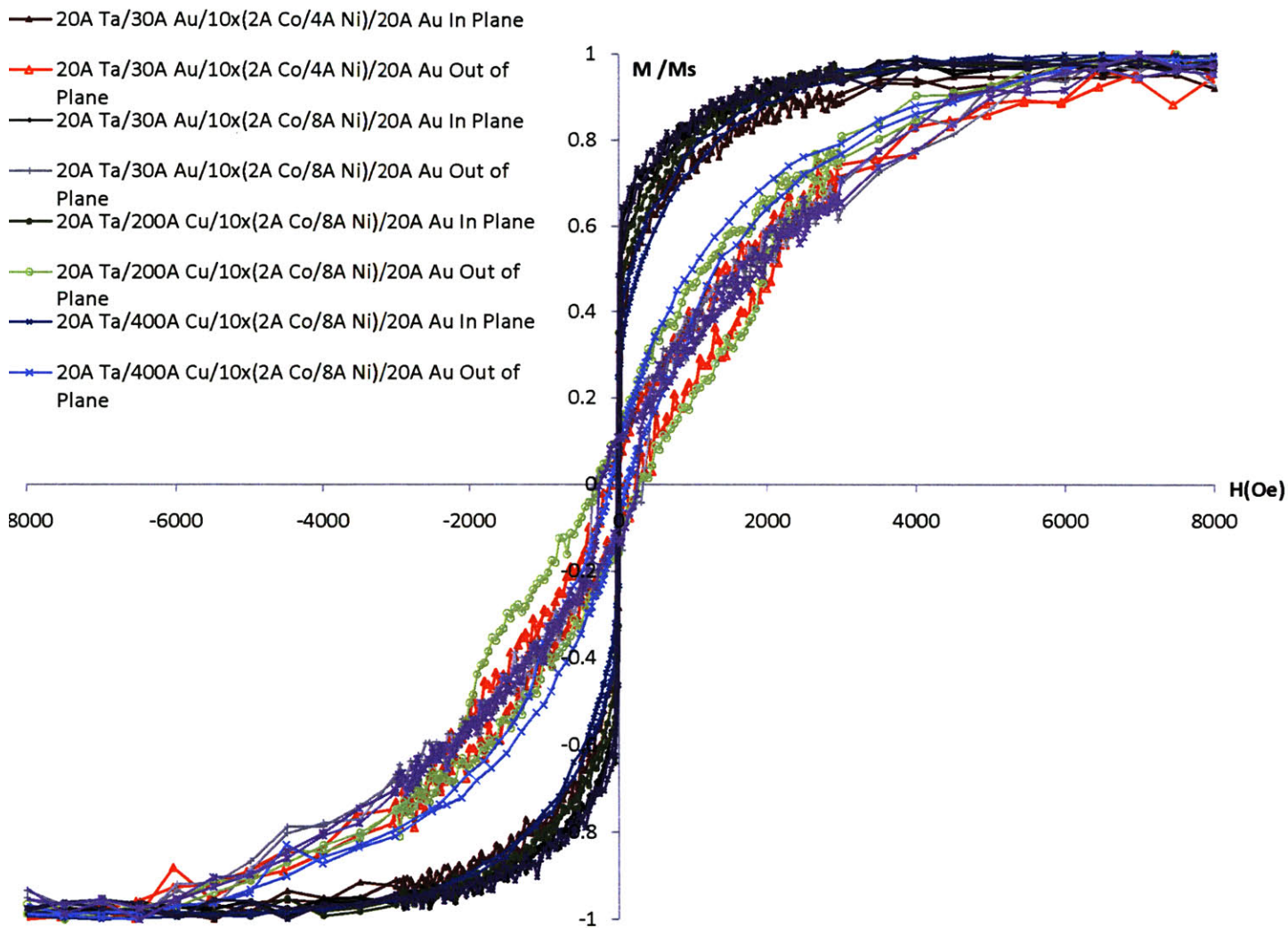


Figure 7: In plane and out of plane magnetic hysteresis loops for various Co/Ni multilayers on gold or copper buffer layers as measured by vibrating sample magnetometry (VSM). The sample compositions are labeled. IP stand for in plane and OOP stands for out of plane.

Chapter 5: Conclusion and Future Work

We deposited Co/Ni multilayers with ion beam sputtering on gold and copper buffer layers of various thicknesses deposited by DC magnetron sputtering and ion beam sputtering respectively. Magnetic measurements with vibrating sample magnetometry and alternating gradient magnetometry revealed that the Co/Ni multilayer did not have perpendicular magnetic anisotropy and instead displaying parallel magnetic anisotropy. This was explained for each multilayer fabricated except one, as determined by characterization with X-ray diffraction and atomic force microscopy, by either a low degree of (1 1 1) texture in the buffer layer and the multilayer or too much roughness in the buffer layer. In addition we proposed that the sputtering process used to deposit the Co/Ni multilayers resulted in diffusion and alloy formation at the interfaces which caused the parallel magnetic anisotropy instead of the perpendicular magnetic anisotropy.

Future work proposed is to perform tunneling electron microscopy on the interfaces of the sample fabricated to determine the structure of interfaces and the thickness of the layers. This would allow us to determine whether or not the sputtering of the multilayers was leading to diffusion and alloy formation at the Co/Ni interfaces. In addition TEM would allow us to confirm that the Co and Ni layer thicknesses were what we intended and would allow us to obtain a better understanding of the roughness of the interfaces. Also, a Co/Ni multilayer with the same structure as those in this study should be deposited with evaporation on a smooth thick gold buffer layer also deposited with evaporation with the same conditions as used in literature for the fabrication Co/Ni multilayers with perpendicular magnetic anisotropy. The resulting multilayer should be magnetically characterized to determine whether or not it has perpendicular magnetic anisotropy which it most likely will as it has been shown to in literature. If this sample does display perpendicular magnetic anisotropy, Co/Ni layers should be deposited onto an identical evaporated buffer layer with IBS. The presence or absence of perpendicular magnetic anisotropy would then allow us to determine whether the lack of perpendicular magnetic anisotropy in the multilayers we fabricated is due to properties of the buffer layer or the properties of the ion-beam deposited Co/Ni layers.

Depending on the results obtained from TEM and from the evaporated buffer layer experiment the fabrication procedure for the Co/Ni multilayers can be adjusted to try to obtain perpendicular magnetic anisotropy. For example if the problem is with the buffer layer the buffer layer can be deposited with evaporation in the future. If perpendicular anisotropy can be obtained future work would be to further characterize the Co/Ni system as well as to fabricate nanoscale devices with it.

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