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Study of anisotropic magnetoresistance of permalloy films

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STUDY OF ANISOTROPIC MAGNETORESISTANCE OF PERMALLOY FILMS

A Thesis

Presented to

**The Faculty of the Department of Materials Engineering
San Jose State University**

**In Partial Fulfillment
of the Requirements for the Degree
Master of Science**

by

Nagamani Nataraj

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ABSTRACT

STUDY OF ANISOTROPIC MAGNETORESISTANCE OF PERMALLOY FILMS

by Nagamani Nataraj

Sputtered permalloy films of composition 80% Ni-20% Fe were deposited at thicknesses ranging from 2-20 nm (thin films) and 100-200 nm (thick films) at room temperature (300K) and low temperature (125K). The films were annealed at various temperatures and times, to study the effects of thickness, deposition temperature, and anneal temperature and time on the magnetoresistive properties of the film.

The results from this study indicate that anisotropic magnetoresistivity (AMR) of thin permalloy films increases with thickness up to a certain thickness. The value of AMR is also higher for low temperature deposited thin films. The above results can be explained by assuming that the surface scattering is more isotropic than the bulk scattering and that scattering at free surfaces (associated with voids) is more isotropic than grain boundary scattering. The operating mechanism for changes in AMR, texture, and coercivity during anneal treatment appears to be grain growth.

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1. INTRODUCTION

1.1 Scope of magnetoresistive effect

The magnetoresistive effect in ferromagnetic materials is very valuable to the magnetic recording industry. Billions of dollars are being spent on the research of magnetoresistive heads^(1,2) as an alternative to the inductive head for use in the disk-drive and recording industry. The inductive read heads are limited to low density media. They run out of sensitivity at densities over 1 Gbytes/sq.inch⁽³⁾. Magnetoresistive heads, on the other hand, can successfully read high density media (~10 Gbytes/sq.inch) because they are able to detect weak signals. Magnetoresistive heads use the magnetoresistive effect for sensing changes in the orientation of the magnetic domains, where a small change in magnetic field produces a very large change in resistance. Magnetoresistive heads, therefore, have better resolution than inductive heads. Magnetoresistive heads cannot replace inductive heads entirely, because they cannot write or erase. In order to achieve optimum writing and reading, dual element heads (inductive for writing and magnetoresistive element for reading) are being investigated.

In this chapter, the scope, background and the past studies conducted on anisotropic magnetoresistance will be reviewed. This will be followed by the objective of the thesis. In Chapter 2, the experimental preparations, parameters, and calculations will be presented. Chapter 3 includes the results and discussion of the experiments. Finally, Chapter 4 will conclude the entire work.

1.2 Theory of electrical conduction

Electrical conduction is phenomenologically different in metals, such as Fe and Ni, as compared to metals, such as Al and Cu, for two main reasons: Fe and Ni are transition metals, and they are ferromagnetic⁽⁴⁾.

In ordinary metals, the electrons can be differentiated as core electrons and valence electrons. Core electrons have a large probability density at the center and a small one at the corners of the Wigner-Seitz cell. As a result, the core electrons do not contribute to the cohesive energy of the solid metal. In terms of a band model, they are housed in fully occupied bands. Valence electrons, on the other hand, are distributed uniformly throughout the crystal. They occupy a broad conduction band, which is nearly parabolic in nature. For a parabolic band with nearly free electrons, the electrical resistivity can be expressed in terms of electronic charge e , an effective electron mass m^* , a relaxation time τ and density of electrons n as

$$\rho = \frac{m^*}{ne^2\tau} \quad \text{Eqn. 1}$$

Transition metals have partially filled d shells. The electrons in the d shell cannot be classified as either core or valence electrons. The d shell electrons differ from valence electrons as their spatial distribution is not uniform. Bands formed from the d electrons are much narrower than a normal conduction band. They are not core electrons either, since the d bands are not filled completely and d electrons have an important probability density in the region between neighboring atoms, indicating that such metals may have high cohesive energies or boiling points.

In noble metals such as Au, Ag, and Cu, there is a d band, and hence a contribution to the cohesive energy, but the d band is completely filled up.

The effect of adding d electrons to a metal on the electrical conduction in which conduction takes place by s -like carriers can be understood using Eqn.1. While addition of d electrons will decrease resistivity by increasing n , the effect is only moderate since m^* is large. On the other hand, transition elements provide d states, and hence the density of states at the Fermi surface increases, and the relaxation time τ decreases. Assuming τ is inversely proportional to the number of states⁽⁴⁾, ρ of transition metals exceeds that of noble metals.

Magnetoresistance, as the name implies, refers to the change in resistance due to the application of a magnetic field. There are two types of magnetoresistance: positive and negative⁽²⁾. Positive magnetoresistance refers to the increase in resistance with increase in the intensity of the applied magnetic field. This increase in resistance is due to Lorentz's forces⁽⁵⁾. This behavior is typically observed in non-ferromagnetic metals. Negative magnetoresistance refers to decrease in resistance with increase in intensity of the applied field. This is a common feature of ferromagnetic materials, such as Ni, Fe, and Co, and this property is made use of in magnetoresistive devices. If θ is the angle between magnetization direction and current direction, then resistance R is proportional to the square of cosine θ . Negative magnetoresistance is commonly accompanied by anisotropic magnetoresistance. Anisotropic magnetoresistivity (AMR), $\Delta\rho$, is the difference between resistivities when the magnetization is parallel to the current versus perpendicular to the current⁽⁵⁾, defined in Eqn. 2.

$$\Delta\rho = \rho_{||} - \rho_{\perp} \quad \text{Eqn. 2}$$

where $\rho_{||}$ is the resistivity when magnetization and current are parallel to each other, $\rho_{\perp}$ is resistivity when magnetization and current are perpendicular to each other.

For ferromagnetic materials, the mean free paths of the upspin and downspin electrons are different^(5,6) in the presence of an external magnetic field. This is understood to be due to the different densities of states for the upspin and downspin d electrons at the Fermi energy⁽⁵⁾. Hence the scattering rates for the upspin and downspin d electrons are different. Thus the effective scattering cross-section that a conduction electron faces in the parallel (when the current and magnetization are parallel to each other) and perpendicular (when the current and magnetization are perpendicular to each other) orientations is different. This is why AMR exists in all ferromagnets. Fig. 1 is a schematic representation of the scattering cross section in the two orientations.

Fig. 1 Schematic representation of the scattering cross-section in the parallel and perpendicular orientations for a conduction electron.

The variation of resistivity with respect to the orientation of the magnetic field is shown in Figure 2. The curves are extrapolated from the saturated regions to zero induction states, B_0 . The parallel and perpendicular resistivities have different values of B_0 .

$$B = H + 4\pi M(1 - D) \quad \text{Eqn. 3}$$

where D is the demagnetizing factor for transverse and longitudinal fields. The curves $\rho_{||}$ and $\rho_{\perp}$ are one way magnetization plots, because the hystereses of the curves are not shown. The double headed arrows A and B indicate regions where the magnetization has not reached saturation in the parallel and perpendicular orientations respectively (saturation is attained when all the magnetic moments line up in the direction of the applied field).

Anisotropic magnetoresistivity ratio (AMR ratio), is defined as $\Delta\rho / \rho_{avg}$, where ρ_{avg} is the average resistivity computed as

$$\rho_{avg} = \frac{1}{3}\rho_{||} + \frac{2}{3}\rho_{\perp} \quad \text{Eqn. 4}$$

For permalloy, anisotropic magnetoresistance is typically about 2% for a 10 nm thick film, and about 4% in the bulk⁽⁷⁾.

Fig.2 Schematic one-way magnetization plot showing how AMR is calculated.(Ref: Campbell andFert)

1.3 Past work

AMR of permalloy films depends on various parameters such as thickness, composition, deposition temperature and anneal treatment. Several articles which discuss parameters of interest to the present study, such as thickness and deposition temperature, will be summarized in this section. The summary of past work will be drawn from more papers.

1.3.1 Thickness dependence

Mitchell *et al.*⁽⁷⁾, described the film thickness dependence of AMR ratio of Fe-Ni films. The films were vacuum deposited onto a heated substrate (573K) in the presence of a constant magnetic field of 20 Oe oriented in the plane of the film and the thicknesses varied from 17.5 to 260 nm. Pressure in the system was on the order of 10^{-6} Torr. Composition and thicknesses were controlled by wrapping an amount of iron wire and nickel wire onto a tungsten filament to give the desired composition and of an amount to yield a predetermined thickness.

The resistance of the films was measured using a four terminal resistance measurement. Thickness was measured by two techniques: multiple beam interferometer for thicker films and by optical transmission density for thinner films.

Figure 3 shows the dependence of AMR on thickness for a 82%Ni - 18% Fe film . It can be seen that while the average resistivity decreases with increase in thickness, the AMR is independent of thickness at thicknesses greater than 10 nm, and thus the AMR ratio increases with thickness. The authors conclude that the average resistivity depends on thickness because films of different thicknesses have different surface area and/or grain

boundaries. Resistivity increases with increase in surface area and grain boundary area. For very thin films, the surface scattering of conduction electrons becomes a significant fraction of the total number of collisions because the thickness is comparable or less than the mean free path of an electron at the temperature concerned, thus the resistivity increases significantly.

Fig. 3 Dependence of anisotropic magnetoresistivity ratio on thickness of the film for 82% Ni - 18% Fe (Ref. Mitchell *et al.*). Crosses are average ρ , circles are AMR ratio, and triangles are AMR.

Yeh *et al.*⁽⁸⁾ studied permalloy films in the thickness range 22 to 549 nm and found oscillatory dependence of AMR on thickness when saturated in a magnetic field. Resistivity and AMR ratio showed oscillatory dependence as well, but to a much lesser degree. The primary source of this oscillatory thickness dependence was attributed by the authors to the quantum size effects arising from the quantization of the electron quasimomentum. Valleta⁽⁹⁾ studied permalloy films ranging from 10 to 120 nm in thickness and found no oscillatory thickness dependence of resistivity, AMR, and AMR ratio.

1.3.2 Substrate temperature dependence

A. Iwaide *et al.*⁽¹⁰⁾ describe the dependence of anisotropic magnetoresistivity ratio on deposition temperature. The material used in this study was 80%Ni - 15%Co - 5%Fe. The samples were deposited by electron beam evaporation in a vacuum of 10^{-6} Torr. The samples were 3.5 mm wide and 15 mm long strips with thicknesses of 100 nm. AMR ratio measurements of samples were each performed after ac demagnetization at 50 Hz. Magnetic fields of up to 200 Oe were applied using a Helmholtz coil. Magnetic properties were measured using a vibrating sample magnetometer (VSM).

Figure 4 shows the results of the experiment. The substrate temperature was varied from 298 to 673K. The average resistivity remained fairly constant up to 473K and decreased slightly with further increase in substrate temperature while the AMR increased. Hence the AMR ratio increases with substrate temperature. The authors conclude by pointing out that the evaporated films at substrate temperatures of 673K yield AMR ratios of 5% or above and summarizing their work. The effect of substrate temperature on the microstructure of the films deposited at temperatures greater than 473K is mainly to increase the grain size. Grain growth is expected due to increase in surface mobility, thus allowing the film to decrease its total energy by growing large grains and thus decreasing the grain-boundary area.

Fig. 4 Dependence of substrate temperature on ρ , $\Delta\rho$, and $\Delta\rho/\rho$ for 80%Ni-15%Co-5%Fe (Ref. Iwaide *et al.*)

1.3.3 Summary of past work

The previous work on the thickness dependence of AMR are inconclusive. Comparing the results from Mitchell *et al.*⁽⁷⁾ and William and Mitchell's⁽¹¹⁾ study, it can be seen that the reported behavior of AMR as a function of thickness is inconsistent. AMR was found to be invariant with thickness in Mitchell *et al.*'s study, whereas AMR was found to increase with thickness in William and Mitchell's study. The films used in the two studies had the same composition, film thicknesses ranging from 7-270 nm, and were deposited by evaporation at pressures of 10^{-6} Torr. The two differences were in the rate of evaporation of the films and the ambient in which the measurements were conducted. The measurements in Mitchell *et al.*'s study were conducted at atmospheric pressure, while the measurements in William's study were conducted in vacuum.

Iwaide *et al.*'s study on the effect of substrate temperature was conducted only on a 100 nm thick film. AMR increased with deposition temperature. The study measured only the electrical and magnetic properties of the film, but structural information was not investigated.

Clearly, a good understanding of the dependence of AMR on thickness is very valuable. In order to correlate the effect of thickness on AMR, more data on films with variations in thickness are needed. A definite relationship between magnetoresistive properties as a function of thickness and temperature, for a given deposition condition, can be established by monitoring the electrical, magnetic, and structural (such as grain size and texture) properties.

1.4 Objective

The objective of the thesis was to study the effect of thickness, deposition temperature, anneal temperature and time on the magnetoresistive properties ρ_{avg} , $\Delta\rho$, and coercivity for sputtered permalloy films of composition 80%Ni-20%Fe, in order to establish a definitive relationship between the magnetoresistive and structural properties.

2. EXPERIMENTAL

In order to accomplish the objective, films of varying thicknesses had to be deposited by sputtering. The first section discusses the design of the masks for the contact pads and the specimen, which is followed by the design of the probe holder. The next section depicts the set-up of the data acquisition system. This is followed by the section describing the deposition of contact pads and film. The last section on experimental measurements and calculations outlines the resistance, coercivity and x-ray measurements and the anneal treatment conducted on the films.

2.1 Experimental preparation

2.1.1 Design of contact pads, specimen and probe holder

The contact pads along with the specimen are shown in Figure 5. Since the dimensions of the specimen are very small, the specimen was deposited on the contact pads, and these pads provided a larger area for the probes to make contact while taking measurements. The contact pads were designed for four probe analysis (two each for current and voltage measurements), and provisions were also made for the measurement of Hall voltage although the Hall measurements were not conducted in this study.

The next step involved the design of the probe holder, with dimensions similar to that of the contact pads. The specimen probe assembly is shown in Figure 6. The probes were gold plated Ni spring loaded pins that were housed in wire wraps. The wire wraps were glued to the probe holder for a snug fit. The probes make contact with the contact pads. The probe holder was made of a material called "ultem" that is a polyamideimide. One end of the wire wrap housed the probe, while the other end was used for providing external

leads for current and voltage. Three Cu-Be strips at 120° to each other were attached to the probe holder along the periphery. The spring action of this material enabled the sample along with the substrate, to be clipped to the probe holder during analysis to provide better contact as well as ease of handling. The probe holder was designed to be suspended between the poles of the magnet. The entire sample-probe assembly could be rotated between the poles of the magnet, enabling anisotropy measurements. This configuration of the probe holder was chosen because it would allow in-vacuum measurements of specimens; however, in the work described here, the samples were removed from vacuum before measurements.

The data acquisition system was automated using a general purpose interface bus (GPIB). The multimeter measured the resistance of the films, while the gaussmeter measured the strength of the magnetic field. Fig. 7 is a schematic of the wiring of the data acquisition system.

Fig. 5 Plan view of the contact pads and the sample

Fig. 6 Specimen probe holder along with the Cu-Be strip

Fig. 7 The wiring diagram of the data acquisition system

2.1.2 Deposition of contact pads

The first step was to deposit the contact pads through the contact masks on the substrate. The quartz substrate was first cleaned in sulfuric acid and rinsed with de-ionized water and then thoroughly dried in a dessicator. The 0.5 inch substrates were held by a substrate holder. Twenty-five substrates were used in each batch. First, 10 nm of chromium was deposited onto the substrate. Chromium was deposited because it can adhere well to both quartz and gold, and gold does not adhere to quartz very well. A vacuum of 5×10^{-6} Torr was maintained in the evaporator. The rate of deposition is controlled by the amount of current supplied. Crystal monitors record the weight of the material deposited. By knowing the density of the material being deposited, the thickness can be computed. Typical deposition rates were 0.1 nm /second. After depositing a thickness of 10 nm of chromium, a thickness of 50 nm of gold was deposited.

2.1.3 Film deposition

After depositing the gold contacts on to the substrate, the sample deposition was performed through the sample mask by sputtering techniques. The sputtering machine, manufactured by SPI (Sputter Films, Inc.) has 5 DC magnetron sources. Argon was the sputtering gas used at a pressure of 5 mTorr. The base pressure in the sputtering chamber was of the order of 10^{-9} Torr. The deposition rates were 0.1 nm /sec. The targets were 2 inches in diameter, 0.25 inch in thickness. The composition of the target was similar to that of the film. In this study, only the composition 80% Ni 20% Fe was investigated. This composition was chosen because it has low magnetostriction and magnetocrystalline anisotropy and good magnetoresistive properties. The substrates were held by a puck, which in turn was placed on a copper plate, so that the substrate temperature could be varied by heating or cooling the Cu plate. By cooling the Cu plate with liquid nitrogen for

four hours prior to deposition, a substrate temperature of 100-125K was achieved. The two substrate temperatures used were 125K and 300K. The substrate temperatures chosen were limited to the choice of the cooling agent used. The thickness was monitored using crystal monitors. Thicknesses investigated were 2, 4, 6, 8, 10, 15, 20, 100, 150, and 200 nm. During deposition, a field of 50 Oe was applied parallel to the plane of the film using a magnet positioned beneath the substrate holder. The direction of the field, which is along the major axis of the film, defines the easy axis of the film. After depositing the sample to the required thickness, a Tantalum overlayer 5 nm thick was applied to prevent the films from oxidizing. Ta was chosen because it is microcrystalline, and forms a self-limiting oxide. Ta does influence the magnetic properties; however, all the samples were subjected to the same degree of influence.

2.2 Experimental measurements and calculations

2.2.1 *Resistance measurements*

The sample was placed in the magnet assembly, as shown in Fig. 8. The magnetic field was incremented from zero to the maximum positive value (typically 15 Oe), which is necessary to saturate the magnetization of the film at that orientation, from the maximum positive value to the minimum negative value (typically -15 Oe), and from minimum negative value back to zero (null). The resistance undergoes a hysteresis with the magnetic field. Resistance is measured in both the parallel (current and magnetization are parallel to each other) and the perpendicular (current and magnetization are perpendicular to each other) orientations. The difference in resistance at saturation between the parallel and perpendicular orientations is the AMR.

Fig. 8 Set-up of the magnet assembly to measure parallel and perpendicular resistivities.

2.2.1.1 Calculation of resistivity

The resistivity can be calculated from the resistance of the film using the following equation

$$\rho = Rt\left(\frac{w}{l}\right) \quad \text{Eqn. 5}$$

where l is the length of the film, w is the width of the film, and t is the thickness of the film. The ratio l/w is also called the aspect ratio, and its calculation is shown in Appendix A. The aspect ratio has a value of 15 for the specimen geometry used in this study.

2.2.1.2 Correction for Ta overlayer

Most films have a 5 nm thick Ta overlayer to protect the films. The Ta and permalloy can be thought of as two resistors in parallel. The measured resistance is a sum of the parallel resistances of the permalloy and Ta films. The corrected permalloy resistance can be obtained using the equation

$$\frac{1}{R_{NiFe}} = \frac{1}{R_{meas}} - \frac{1}{R_{Ta}} \quad \text{Eqn. 6}$$

Room Temp deposition: The room resistivity of Ta was known to be $200 \mu\Omega \text{ cm}^{(23)}$. From this value, the resistance of the Ta overlayer was calculated to be 6000Ω using Eqn 5, where t , the thickness of Ta, is 5 nm.

Low temperature deposition: On a quartz substrate with contact pads, 5 nm of Ta was deposited directly on glass through the sample mask, and the measured value of low temperature deposited Ta resistivity, which was $358 \mu\Omega \text{ cm}$, was used for correcting the low temperature deposited samples. The parallel resistivity of Ta was subtracted from the measured value to get the resistivity of permalloy, just as outlined in the case of room temperature deposited samples above.

2.2.2 Calculation of coercivity

Fig. 9a is a plot of the measured film resistance in the parallel orientation (current and magnetization are parallel to each other) as a function of magnetic field. Fig. 9b is a plot of the measured film resistance in the perpendicular orientation (current and magnetization are perpendicular to each other) as a function of magnetic field. Coercivity can be interpreted as the intrinsic field which opposes the change in the direction of the external applied field. Coercivity values can be obtained from resistance vs field measurements. The averaged horizontal distance between the two maxima (in the perpendicular orientation, as shown in Fig. 9b) and minima (in the parallel orientation, as shown in Fig. 9a), which represent the maximum change in resistance, is the coercivity. Sometimes the peaks were not very well defined due to magnetic field fluctuations. Judiciousness had to be used to determine where the peak maxima (or minima) were. Coercivity was calculated manually, and may have substantial experimental error.

Fig. 9 Calculation of coercivity from the measured value of resistance for (a) current parallel to field and (b) current perpendicular to field. The insets show the ideal B-H loop which would be measured in the same orientation.

2.2.3 X-ray diffraction

X-ray analysis of thin films does not produce a strong recordable diffraction pattern because of insufficient scattering power of an atom for an x-ray. As a result, two sets of thick films were deposited. One set of room temperature deposited films had thicknesses of 100, 150 and 200 nm, respectively. Another set of low temperature deposited films each 100 nm thick, was deposited. Both sets of films were subjected to isochronal anneal similar to the low temperature deposited thin films, and x-ray analysis was done prior to and after each anneal. X-ray analysis provided information on the grain size, texture, and strain in the films.

Calculation of grain size

The diffractometer used was a rotating-anode tube using a monochromatic $\text{CuK}\alpha$ radiation. The average thickness of the [111] grains, t , was calculated using Scherrer's⁽¹³⁾ formula, which is given below

$$t = \frac{\lambda}{B \cos \theta_B} \quad \text{Eqn. 7}$$

where λ is the incident wavelength of the x-ray, B is full-width half-maximum of the [111] diffracted peak in radians, and θ_B is the Bragg angle.

Texturing

The ratio of the integrated intensity in the [111] direction to that in the [200] direction gives an indication of the degree of texturing. If the ratio is larger than that of the value of a bulk polycrystalline 80%Ni 20%Fe sample, which is 1.25⁽¹⁴⁾, then the films are textured. The integrated intensities can be obtained directly from the plotting program.

Lattice strain

Lattice strain is calculated by noting the percentage difference between the values of planar spacing of an ideal 80%Ni 20%Fe sample and the ones obtained experimentally according to the relation

$$100\left(\frac{d_{meas} - d_{ideal}}{d_{ideal}}\right) = \%strain \quad \text{Eqn. 8}$$

The d-spacing of the sample was compared to the d-spacing of an isotropic material, and the percentage difference was used to obtain the strain in the lattice.

2.2.4 Anneal treatment

The films were annealed to find the effect of anneal treatment (such as reduction in defect density, grain growth, coalescence, and strain) on the magnetoresistive properties (ρ , $\Delta\rho$, $\Delta\rho / \rho$) and coercivity. Annealing also provides information on the relationship between structural changes and property changes.

The films were annealed in a well insulated furnace in a reducing atmosphere (4.1% hydrogen, balance nitrogen). Table 1 indicates the temperatures and times at which low temperature and room temperature films were annealed. The room temperature deposited thin films (<20 nm) were annealed at two temperatures. One set was annealed at 650K for 4 hours, while the other was annealed at 550K for 24 hours. The low temperature deposited thin films were subjected to isochronal anneal at 500, 550 and 600K for 13 hours. After the isochronal anneal, the films were further annealed at 600K for additional 13 hour periods, until the films were no longer electrically intact (or mechanically intact).

Table 1 Anneal treatment for room temperature and low temperature deposited permalloy films

Film type	500K/13h	550K/13h	550K/24h	600K/13h	600K/26h	650K/4h
RT film			Set A			Set B
LT film	Set C	Set C		Set C	Set C	

3. RESULTS AND DISCUSSION

There are three sections in this chapter. One section describes the effect of Ta overlayer, the second describes the as-deposited characteristics of the films as a function of deposition temperature, and the third section discusses changes in film properties after annealing.

3.1 Effect of Ta overlayer on room temperature deposited films

In this section, evidence will be presented to show that Ta suppresses anisotropic magnetoresistivity when used as an overlayer. This will be shown by comparing the $\Delta\rho$ of the room temperature deposited Ta coated and uncoated permalloy films. Corrections for the Ta overlayer were made by subtracting the parallel component of Ta resistance from the measured resistance, using Eqn. 6, to get the corrected resistance of the permalloy film. The corrected film resistance is converted into resistivity using Eqn.5. The uncoated films are reported "as-measured" without any correction.

Fig. 10 is a plot of corrected resistivity of permalloy vs permalloy thickness for Ta coated and uncoated films. There is very little difference in resistivity between the coated and uncoated films, implying that both films may be oxidized to the same extent.

Another observation that can be made is that the resistivity of the 2 nm thick film increases by almost an order of magnitude from that of the 4 nm film. This suggests that the 2 nm film may be discontinuous. As a result, the 2 nm thick films were not used in the experiments in this study. Also, Fig. 10 indicates that the bulk permalloy value, $30\ \mu\Omega\ \text{cm}$,

is not reached. Thus, film thicknesses greater than 20 nm are necessary to attain bulk properties.

Fig. 10 Corrected permalloy resistivity vs permalloy thickness for Ta coated and uncoated films.

Resistivity decreases with increase in thickness (see Fig. 10) and this may be due to various reasons. Scattering of conduction electrons by surfaces becomes more dominant in thin films. Thin films may have smaller grains (and hence larger grain boundary area and hence increased grain boundary scattering) than thick films. Resistivity is also dependent on point defect density in films, which may systematically vary as a function of film thickness.

Fig. 11 is a plot of anisotropic magnetoresistivity vs thickness for the coated and uncoated films. Two prominent observations can be made from Fig. 11: $\Delta\rho$ increases with thickness in both cases and the uncoated films have a higher $\Delta\rho$ than coated films.

The increase of $\Delta\rho$ with thickness can be explained by assuming that the scattering at the surface and near surface regions is more isotropic than the bulk. At small thicknesses, there is a large surface-to-volume ratio, and the scattering is more isotropic. As the thickness increases, the surface area to volume ratio decreases, and the scattering is less isotropic, and $\Delta\rho$ increases. Uncoated films have a higher $\Delta\rho$ than coated films. This can be explained if Ta suppresses $\Delta\rho$ by suppressing magnetization. Thus, Ta is not a good choice for an overlayer because it suppresses $\Delta\rho$.

Fig. 11 Anisotropic magnetoresistivity of permalloy films as a function of permalloy thickness for Ta coated and uncoated films

3.2 Effect of deposition temperature

3.2.1 *Structural aspects*

The first part of this section discusses the continuous film concept model. This will be followed by a section that discusses the effect of deposition temperature on structural properties such as grain size, texture, and strain. It must be pointed out that the structural analysis, using x-ray data, was performed only on thick films (films 100-200 nm thick).

3.2.1.1 *The continuous film concept as a function of thickness*

In general, there are four stages of film growth⁽¹⁶⁾: the island stage, the coalescence stage, the channel stage, and the continuous film, as shown schematically in Fig. 12.

In the island stage, the condensate species form nuclei on impingement with the substrate. The size of the nuclei formed at a given substrate temperature is dependent on the vapor pressure of the condensate species. High boiling point, and hence high vapor pressure species such as Ni and Ta, form smaller nuclei than low vapor pressure species. The nuclei grow to form islands. The size of the island formed depends on the adatom mobility which is a function of substrate temperature. Low temperature films are expected to form smaller islands since the adatom mobility at low temperatures is low. Room temperature films, on the other hand, are expected to form bigger islands due to higher adatom mobility.

In the coalescence stage, the islands coalesce and cover larger areas of uncovered substrate resulting in secondary nuclei formation between islands.

In the channel stage, the islands become elongated and join to form a continuous network structure in which the deposited material is separated by long, irregular, and narrow

channels of width 5 to 20 nm. As deposition continues, secondary nucleation occurs in these channels, and the nuclei are incorporated into the bulk of the film as they grow and touch the sides of the channel. At the same time, the channels are bridged at some points and fill in rapidly in a liquid-like manner.

Eventually, most channels are eliminated and the film is continuous except for some irregular holes. Secondary nucleation takes place on the substrate within these holes, and the growing nuclei are incorporated (in a liquid-like manner) into the continuous regions of the deposit. The holes contain many secondary nuclei, which form secondary islands by coalescing, and eventually fill the holes.

In the continuous film stage, considerable changes in the orientation of the islands occur during the growth of the film. Some recrystallization may occur during coalescence, so that the grain size of the completed deposit film is large compared with the average separation of the initial nuclei.

In general, the island stage occurs in films about 2-3 nm thick, the coalescence stage in 3-20 nm thick films, the network stage in 20-35 nm thick films, the hole stage in 35-45 nm thick films, and the continuous film occurs in films over 45 nm thick for a noble metal like Au which has a tendency to ball-up rather than form a continuous film⁽¹⁶⁾. Nickel and iron, however, may form a continuous film at smaller thicknesses.

Schematic of the stages of film growth.

Schematic of the shape changes during coalescence.

Fig. 12 Schematic of the continuous film concept model

3.2.1.2 Dependence of grain size on deposition temperature

Fig. 13 is a plot of grain size vs thickness for room temperature and low temperature deposited films obtained from x-ray diffraction data. There are two main observations that can be drawn from this figure. For a given thickness, such as 100 nm, the grain size of the room temperature deposited film is 33% larger than the low temperature deposited film. This can be explained using the continuous film concept model explained in Section 3.2.1.1. The size of the grain formed depends on the adatom mobility which in turn depends on the deposition temperature. The room temperature films thus have bigger grains at a given thickness due to increased adatom mobility.

Another observation that can be made from Fig. 13 is that the grain size increases with thickness for a given deposition temperature. This can also be understood using the continuous film concept model. The thicker films, being farther along in the deposition process, can coalesce and form bigger grains compared to a thinner film at a given temperature.

Fig.13 Grain size vs thickness as a function of deposition temperature

3.2.1.3 Dependence of texture on deposition temperature

Fig. 14 is a plot of texture vs thickness for room temperature and low temperature deposited permalloy films. The texture in bulk polycrystalline films is nominally 1.25⁽¹⁴⁾. There are two observations that can be made from Fig. 14. Both room temperature and low temperature deposited films are highly textured. Although the effect of deposition temperature on texture is not very apparent from the data, it can be clearly seen that at least for the room temperature films, texture increases with thickness. This is consistent with the grain size dependence, as explained in Section 3.2.1.2. Thicker films have bigger grains because they have had more time to coalesce and grow compared to the grains of the thinner films. Texture is a measure of the ratio of I[111] to I[200], and reflects the increase in size of the (111) and (200) oriented grains. Increase in texture could be brought about by more ways than one. Either I[111] alone could increase, or I[200] alone could decrease, or I[111] could relatively increase over I[200].

Fig. 14 Texture as a function of thickness for room temperature and low temperature deposited films.

3.2.1.4 Dependence of strain on deposition temperature

Fig. 15 is a plot of strain in the [111] direction for as deposited room temperature and low temperature films 100-200 nm thick. The strain data shown in Fig. 15 represents the strain that is normal to the plane of the film. Strain in films is due to the sum of intrinsic strain which arises during deposition, and thermal strain due to mismatch in the coefficients of expansion of the Si substrate and the permalloy film. The thermal strain on the low temperature as deposited film is compressive in the plane of the film because the permalloy film is constrained by the substrate from expanding when the films are brought to room temperature after deposition.

Fig. 15 % Strain as a function of thickness for room temperature and low temperature deposited films.

At low temperatures, there could also be a large incorporation of contaminants⁽¹⁷⁾ on the substrate and film. If there is Ar incorporated in the substitutional sites at low temperature, the lattice would collapse around the Ar atom isotropically, and there would exist an isotropic intrinsic compressive stress. This is because the Ar atoms, with atomic radii 0.088 nm, are very much smaller than the Fe or Ni atoms, with radii 0.172 nm and 0.162 nm, respectively. The total strain is the sum of the thermal, which is compressive in plane, and intrinsic, which is isotropically compressive, strains. The measured strain is compressive in the perpendicular plane, or tensile in-plane, which is consistent with the expected behavior.

Room temperature samples may have less incorporation of Ar than low temperature films. The thermal strain, due to the mismatch in the thermal expansion coefficient between the quartz substrate and permalloy film, may be slightly tensile. This is because the substrate temperature increases a few degrees during sputtering. The overall strain in the as deposited condition for room temperature samples is found to be slightly compressive, which is consistent with the expected result.

3.2.2. *Electrical properties*

In this section, the properties of the room temperature and low temperature deposited films will be compared.

3.2.2.1 *Resistivity as a function of deposition temperature*

The resistivity of the as deposited room temperature and low temperature films as a function of thickness is shown in Fig. 16. Both room temperature and low temperature films have been corrected for the Ta overlayer. From Fig. 16 it can be seen that there is very little difference in resistivities between the two sets of films, implying that resistivity is not influenced by deposition temperature in the thickness range 4-20 nm. For both deposition temperatures, resistivity decreases as thickness increases, as discussed in Section 3.1.

Besides grain morphology, resistivity could also be affected by point defect density, such as vacancy concentration, and incorporation of gases in films. Point defect concentration may not be significantly different at the two temperatures under consideration, while incorporation of gases in films may be higher at low temperatures than at room temperatures. The increase in either of the two factors would increase the resistivity of the films. Since there is very little difference in the resistivity of the room temperature and low temperature deposited films, the point defect density and gas incorporation do not seem to play a major role in determining the resistivity of the films.

Fig. 16 Resistivity as a function of thickness for room temperature and low temperature deposited films

3.2.2.2 Anisotropic magnetoresistivity as a function of deposition temperature

$\Delta\rho$ as a function of thickness for the room temperature and low temperature films is presented in Fig. 17. $\Delta\rho$ increases with thickness only up to 10 nm. This can be explained by assuming scattering at the surface is more isotropic than scattering in the bulk, as discussed in Section 3.1. There are three main observations that can be made from Fig. 17. At thicknesses below 6 nm, $\Delta\rho$ is independent of deposition temperature. $\Delta\rho$ is dependent on thickness only up to 10 nm; above that, $\Delta\rho$ is independent of thickness. This implies that a continuous film is formed at a thickness of 10 nm.

Fig. 17 $\Delta\rho$ as a function of thickness for room temperature and low temperature deposited films.

At thicknesses greater than 10 nm, the low temperature films have a higher $\Delta\rho$ than room temperature films. A plausible explanation is that low temperature films form a continuous film at a smaller thickness than room temperature films for a given set of conditions. Hence, the low temperature films, at a given thickness, may have smaller grains, and hence more grain boundary area. The room temperature films, on the other hand, may have larger grains at the same thickness, and hence smaller grain boundary area but more voids and hence more free surface associated with the voids. If the scattering at the voids/free surface is more isotropic than the grain boundary scattering, then this could explain why room temperature films have smaller values of $\Delta\rho$ than low temperature films.

A low temperature film has smaller grains, and hence more grain boundary area. Although the effect of grain boundary scattering on AMR is not clearly understood, it is plausible that between the adjacent grains, the direction of magnetization is continuous (not very discontinuous) and a single domain configuration exists. A room temperature film, on the other hand, has larger grains and smaller grain boundary area. Voids are formed between large grains (due to packing of large grains). The voids and hence the free surface associated with them is larger in room temperature films. The direction of magnetization between two adjacent grains may be discontinuous and more than one domain may exist. This would imply that the AMR would be more anisotropic in grain boundary scattering compared to free surface/void scattering, as observed in Fig. 17.

3.2.2.3 Coercivity as a function of deposition temperature

Coercivity as a function of thickness for room temperature and low temperature deposited films is plotted in Fig. 18. Although there is a large scatter in the data of coercivity vs thickness due to the fluctuations in the magnetic field during measurement, there is no apparent dependence of coercivity on thickness. Low temperature films seem to have a lower value of coercivity than room temperature films. At low deposition temperatures, the magnetic moments may be aligned more uniformly, and the film may have a well defined easy axis compared to the room temperature deposited films. Also, using the argument used in Section 3.2.2.2, the low temperature films, with smaller grains have a continuous domain between two grains. This may also result in a low coercivity value. The room temperature films with bigger grains, and lots of free surface associated with voids, have more domains between two grains. This may result in a high coercivity value, as shown in Fig. 18.

Fig. 18 Coercivity vs thickness for room temperature and low temperature deposited films

3.3 Effect of anneal treatment

The deposited permalloy films were annealed according to the schedule discussed in Section 2.2.4. During annealing, three phenomena are most likely to occur. These are strain relief, reduction in defect density, and grain coalescence or grain growth. If any of these phenomena occur, changes in the electrical and magnetic properties should be observed. Also, microstructural changes may be observed indirectly by x-ray diffraction. In this section, observations of changes in measured properties will be reported and discussed.

3.3.1 *Changes in structure as a function of anneal treatment*

3.3.1.1 *Grain size as a function of anneal treatment*

From the x-ray data on films ranging from 100 to 200 nm in thickness, grain size of the films was calculated (using Eqn. 7) as a function of anneal treatment as shown in Fig. 19. In Eqn. 7, the peak broadening can be due to grain size, strain, or lattice defects, and it is difficult to separate the effect of each of the single factors. There are two main observations that can be made from Fig. 19. The 150 nm film has a grain size increase only after the first anneal where it reaches a saturation size of 33 nm. The room temperature 100 nm film experiences an increase in grain size and attains a size of 33 nm in the first two anneal treatments, after which there is no substantial growth with anneal treatment. The low temperature films experience an increase in grain size with each anneal treatment and stop increasing after a grain size of 33 nm is reached.

This can be envisioned using the schematic shown in Fig. 20. The low temperature films have smaller grains to begin with. During anneal treatment, some grains grow at the expense of others, and these grains increase in size until they grow to a certain extent (33 nm in this case). The driving force, which is reduction in surface area, may be both thermodynamically possible and the kinetics may be favorable. After the grain size reaches 33 nm, however, further increase in grain size may not be kinetically feasible and the grains stop growing. The room temperature films have bigger grains to begin with, and they reach the optimum grain size after the first anneal. At this point, no reasonable explanation for the 33 nm grain size limitation can be provided.

Grain growth has also been observed after low temperature anneal treatments by Butherus and Nakahara⁽¹⁸⁾. Fig. 21 shows TEM micrographs of permalloy films, 50 nm thick, before and after anneal at 670K for 30 minutes from this study. The authors monitored only the resistivity with each anneal treatment. From the TEM micrograph, the grain size of the 50 nm thick films appears to be about 30 nm.

Fig. 19 Average grain size vs anneal treatment for permalloy films of thickness 100-200 nm. 1 = as-deposited. 2 = After 500K anneal for 13 hours. 3 = After 550K anneal for 13 hours. 4 = After 600K anneal for 13 hours. 5 = After 600K anneal for 26 hours.

Fig. 20 Schematic of change in grain size as a function of anneal treatment for low temperature and room temperature deposited films

Fig. 21 TEM micrographs showing the grain structure of 50 nm thick permalloy films. (a) As- deposited. (b) Annealed at 670K for 30 minutes. (Ref. Butherus and Nakahara)

3.3.1.2 Texture as a function of anneal treatment

The effect of texture on anneal treatment differs for the room temperature and low temperature films, as shown in Fig. 22. Room temperature films experience a slight increase in texture after the first anneal, which is consistent with the grain size increase. The low temperature films experience a decrease in texture with each anneal treatment, even though the [111] grains grew with each anneal treatment. This implies that in the low temperature films, both [111] and [200] grains grew, and that the [200] grains grew preferentially to the [111] grains. This experimental observation is not clearly understood.

Narishinge *et al.*⁽¹⁵⁾ found that if the nitrogen was present along with Ar during sputtering of permalloy films, (002) textured grains grew. The lattice strain increases when nitrogen atoms occupy interstitial sites. The increase in elastic energy becomes smaller when nitrogen atoms occupy sites in the (002) planes rather than the (111) planes, and (002) textured grains grow in the films. The lattice constant of (002) textured grains was found to be larger than that of (111) textured grains. They interpret the (002) textured grains to be body-centered tetragonal, in which one crystal axis of the face centered cubic, the *c* axis, is elongated. When nitrogen partial pressure increases, the lattice constant of FCC (111) textured grains becomes smaller, and the lattice constant of *c* axis of BCT (002) textured grains becomes larger. The decrease in texture with anneal treatment in the low temperature films used in this study can be explained if the above argument is valid. Texture decreases with anneal treatment, as more nitrogen interstitials are incorporated from the ambient during annealing. Room temperature films, on the other hand, experience an increase with anneal treatment, and the above argument is not valid.

Fig. 22 $I[111]/I[200]$ as a function of anneal treatment for permalloy films of thickness 100-200 nm. 1 = as-deposited. 2 = After 500K anneal for 13 hours. 3 = After 550K anneal for 13 hours. 4 = After 600K anneal for 13 hours. 5 = After 600K anneal for 26 hours.

3.3.1.3 *Strain as a function of anneal treatment*

Fig. 23 is a plot of percentage strain vs anneal treatment. The strain measurements in the [111] direction measures the change in d spacing perpendicular to the film. The first anneal treatment at 500K relieves the strain in low temperature films. This may be due two reasons: liberation of the Ar substitutionals incorporated during low temperature deposition, and due to thermal strain. The thermal strain is tensile in nature in the plane of the film, as relaxation occurs during anneal and the Si substrate cannot keep up with the metal film. An in-plane tensile thermal strain would result in a compressive measured strain, as observed in Fig. 23. With subsequent annealing, the measured strain gets progressively compressive due to thermal strain.

As shown in Fig.23, the measured strain gets more compressive with each successive anneal in the room temperature films. The thermal strain due to the anneal treatment is tensile in nature in the plane of the film. The measurements indicate compressive strain, because they are made perpendicular to the plane of the film. With subsequent annealing, both room temperature and low temperature films have similar values of strain.

Fig. 23 % Strain as a function of anneal treatment for permalloy films of thickness 100-200 nm measured in the [111] direction. 1 = as-deposited. 2 = After 500K anneal for 13 hours. 3 = After 550K anneal for 13 hours. 4 = After 600K anneal for 13 hours. 5 = After 600K anneal for 26 hours.

3.3.2 Change in electrical properties of films as a function of anneal treatment

3.3.2.1 Resistivity as a function of anneal treatment

Fig. 24 is a plot of resistivity vs anneal treatment for low temperature as well as room temperature deposited samples. From the observed results on 4-20 nm thick samples, it can be seen that the resistivity of films 6 nm and above does not change with anneal treatment, while the 4 nm sample has an increase in resistivity with anneal treatment. Treatment # 1-5 pertain to low temperature films and treatment #6 pertains to room temperature films. To distinguish between the two sets of films, filled data points are used for room temperature films. Even though the low temperature films had an increase in grain size with anneal treatment, there is no change in resistivity in films 6-20 nm thick with anneal treatment, implying thin films may behave differently from thick films. It should be noted that treatment # 3 , which represents low temperature films annealed at 550K for 13 hours, can be compared with treatment #6, which represents room temperature films annealed at 550K for 24 hours. The resistivities of films 6 nm and above in treatment #3 is comparable to that of treatment #6, indicating similar behavior in both room temperature and low temperature films.

The resistivity of the 4 nm film increases with anneal treatment for both room temperature and low temperature deposited films. This can best be explained by assuming grain growth or coalescence to be the dominant mechanism for change in resistivity with anneal treatment. In very thin films, grain growth can reduce connectivity, thereby increasing the resistivity. Also, the resistivity of the room temperature 4 nm film is smaller than that of the low temperature 4 nm film. If the low temperature films form small clusters at deposition compared to room temperature films, they will experience a greater amount of

discontinuity due to grain growth compared to the room temperature film resulting in a higher value of resistivity as observed in the experiment. This is shown in Figure 25.

Fig. 24 Resistivity vs anneal treatment. 1 = as-deposited. 2 = after 500K, 13 hour anneal. 3 = after 550K 13 hour anneal. 4= after 600K 13 hour anneal. 5 = after 600K 26 hour anneal. 6 = after 550K 24 hour anneal*. Treatment #6 (filled symbols) is room temperature film data, while treatments #1-#5 (hollow symbols) represent low temperature film data.

The low temperature 4 nm film may have lots of free surface in the as-deposited condition, as shown in case (a) of Fig. 25. After the first anneal, the low temperature film opens up even more as shown in case (b). With subsequent annealing, the grains start coalescing and this is depicted by case (c). After the last anneal, grains in the low temperature film coalesce to a greater extent, as shown in case (d). Room temperature 4 nm film, on the other hand, does not have nearly as many free surfaces as in low temperature films, as shown in case (e). After the first anneal, the grains coalesce, as shown in case (f), and experience no further change with anneal treatment due to thermodynamic or kinetic factors.

Thin films behave differently from thick films. Since resistivity measurements were made on thin films (<20 nm), and grain size calculations were made on thick films, it is difficult to correlate the resistivity changes with microstructural changes.

Fig. 25 Schematic of plan view of grains in 4 nm film (a) low temp. as deposited film with lots of free surface. (b) low temp. film after anneal opens up even more. (c) low temp. film starts to close with further annealing. (d) low temp. film closes even more at the end of anneal treatment. (e) room temp as deposited film not nearly as much free surface as in low temp. (f) room temp. film after anneal does not change further.

3.3.2.2 $\Delta\rho$ as a function of anneal treatment

Fig. 26 is a plot of $\Delta\rho$ as a function of anneal treatment. All thicknesses 6 nm and above have increase in $\Delta\rho$ after the first anneal after which $\Delta\rho$ stabilizes. The value at which $\Delta\rho$ saturates depends on the thickness of the film. The 15 nm film saturates at $0.6 \mu\Omega \text{ cm}$ which is close to the bulk value. The 6 nm film saturates at a smaller value of $\Delta\rho$, $0.5 \mu\Omega \text{ cm}$. $\Delta\rho$ in the 4 nm film increases with anneal treatment because 4 nm films undergo grain growth / coalescence which increases $\Delta\rho$ according to the assumption.

According to the standard anisotropic magnetoresistivity theory⁽⁵⁾, point defect scatter centers do not affect $\Delta\rho$, since the parallel and perpendicular resistivities are affected the same way. This implies that a mechanism other than point defect annihilation must be responsible for the observed behavior of $\Delta\rho$ with anneal treatment.

The third possible mechanism occurring during annealing is strain relief. However, the parallel and perpendicular resistivities should be affected similarly by strain relief. It is unclear how strain affects $\Delta\rho$.

Of the three mechanisms discussed, coalescence coupled with grain growth offers the best explanation for the observed results.

Fig.26 Anisotropic magnetoresistivity as a function of anneal treatment. 1 = as deposited condition, 2=after 500K 13hour anneal, 3 = after 550K 13 hour anneal, 4 = after 600K 13 hour anneal, 5 = after 600K 26 hour anneal, 6 = after 550K 24 hour anneal. Treatment #6 (filled symbols) refers to room temperature films, while the #1-5 (hollow symbols) refer to low temperature films.

3.3.2.3 *Coercivity as a function of anneal treatment*

Fig. 27 depicts the change in coercivity as a function of anneal treatment for the room temperature and low temperature deposited films. Two main observations can be made from Fig. 27. Annealing treatment at temperatures 600K and below had no apparent influence on coercivity in film thicknesses 4-20 nm, as shown in treatment # 2-5. The same result was obtained when room temperature films were annealed at 550K for 24 hours, indicating that the deposition temperature does not influence the coercivity behavior. However, films 100-200 nm thick, subjected to the same anneal treatment, had a huge jump in coercivity (coercivity increased from 2.0 Oe in the as-deposited condition to 80-100 Oe after the isochronal anneal treatment). The films also had an isotropic hysteresis curve. This clearly indicates that the increase in coercivity that a film experiences for a given anneal treatment depends on its thickness. Thick films experience a large increase in coercivity while coercivity of thin films is not affected by anneal treatment at temperatures 600K and below.

At temperature of 650K, the coercivity of the room temperature deposited thin films 20 nm and 15 nm thick increased ten-fold, while the 4 nm film merely doubled its coercivity value, as shown in treatment # 6 in Fig. 27. Films annealed at temperatures below 600K did not have the sharp increase in coercivity. Miyazaki and Oikawa⁽¹⁹⁾ found that coercivity did not change with anneal temperature up to 700K, above which it increased sharply, and this is shown in Fig. 28. The sharp increase in coercivity was also accompanied by an isotropic hysteresis curve, consistent with the results from the present study. This implies that the easy axis of the films was lost due to annealing.

Fig. 27 Coercivity as a function of anneal treatment for permalloy films of thickness 4-20 nm. 1 = as-deposited. 2 = After 500K anneal for 13 hours. 3 = After 550K anneal for 13 hours. 4 = After 600K anneal for 13 hours. 5 = After 600K anneal for 26 hours. 6 = After 650K anneal for four hours. Treatment # 6 (filled symbols) pertains to room temperature films while #1-5 (hollow symbols) pertains to low temperature films.

Fig. 28 Results from Miyazaki and Oikawa's study showing the extreme temperature at which coercivity increases sharply.

Anderson *et al.*⁽²⁰⁾ found a sharp increase in coercivity when annealed in the temperature range of 573-623K in the presence of a magnetic field parallel to the easy axis of the films. This increase was attributed to grain growth, Bloch wall pinning due to the formation of impurity compounds such as sulfates, and surface coarsening caused by stress relief mechanisms operating in these films. In films 4-20 nm thick, the thickness is too small for the existence of Bloch walls and the observed increase in coercivity could be due to the other possibilities listed above. However, in films 100-200 nm thick, Bloch wall pinning is possible.

4. CONCLUSIONS AND FUTURE WORK

Studies conducted on Ta coated permalloy films and uncoated permalloy films indicate that both films had similar resistivities, implying the same degree of oxidation. Ta also suppresses magnetization. Thus Ta is not a good choice for use as an overlayer.

ρ is independent of deposition temperature for the two temperatures considered and dependent upon thickness in the thickness range studied (4-20 nm). AMR is independent of deposition temperature up to 6 nm, and independent of thickness above 10 nm. It is rationalized that low temperature deposition yields fine grained films with fewer voids, thus having more grain boundary area than room temperature films. Room temperature films, on the other hand, have bigger grains, and more voids between grains, and hence more free surface associated with voids. Scattering at surface and near surface regions may be more isotropic than the bulk. AMR for low temperature films is greater than that for room temperature films. The AMR results of the low temperature and room temperature films can be explained if grain boundary scattering is less isotropic than void / free surface scattering.

Grain coalescence and growth appear to be the dominant mechanisms during deposition and annealing, respectively. The degree of grain growth may be dominated by kinetic factors. Several investigators have attributed the increase in coercivity to grain growth, surface coarsening, and Bloch wall pinning due to the formation of impurities such as sulfates. The observed increase in coercivity in thick films could be a consequence of any one or all of these factors. The temperature at which coercivity increases sharply depends on the thickness of the film.

Future work in this area should focus on understanding the relationship between magnetoresistive properties and structure to a greater detail. The electrical measurements in this study were conducted on thin films (4-20 nm thick), while the structural measurements were conducted on thick films (100-200 nm thick), and it is difficult to correlate the two sets of films. The future study should acquire electrical and magnetic, as well as structural, information from the same set of films, so that the magnetoresistive behavior of the films can be exactly correlated to its microstructure. Although the films used in this study were deposited in ultra-high vacuum, the measurements were conducted at atmospheric pressure and the anneal treatment was conducted under a reducing atmosphere at atmospheric pressure. To avoid the influence of the ambient during measurements and a reducing medium during anneal treatment, measurements and anneal treatment should be conducted in vacuum.

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Appendix

$$\text{ASPECT RATIO} = \frac{L}{W} = \frac{0.1875''}{0.0125''} = 15$$