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# PERPENDICULAR RESISTANCE AND MAGNETORESISTANCE OF Co/Ag MULTILAYERS

By

Shang-Fan Lee

#### A DISSERTATION

Submitted to Michigan State University in partial fulfillment of the requirements for the degree of

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#### ABSTRACT

# PERPENDICULAR RESISTANCE AND MAGNETORESISTANCE OF Co/Ag MULTILAYERS

By

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The resistances of certain synthetic multilayer thin films composed of alternating layers of ferromagnetic (F) and non-magnetic (N) metals decrease significantly with increasing magnetic field — called Giant Magnetoresistance (MR). Both the F layers and the F/N interfaces serve as spin-dependent scattering sources of conduction elections, and a fundamental question involves the relative importance of bulk and interface scattering in giant MR.

The MR is usually measured with the Current flowing In the layer Planes (CIP geometry). The resulting sample resistance ( $\simeq 0.01\Omega$ ) needs only standard measuring techniques, but it is difficult to separate the bulk and interface contributions to the MR, because the currents are mixed by transmission of electrons across the interfaces. In contrast, the MR measured with the Current flowing Perpendicular to the layer Planes (CPP geometry) is harder to measure because a thin film has a small resistance ( $\simeq 10^{-7}\Omega$ ) in the CPP geometry if it is not specially shaped with modern lithography

techniques. But the separation of bulk from interface scattering should be more straightforward, since the current passes through the individual layers and interfaces sequentially.

In this dissertation we show how to simultaneously measure the CPP and CIP MRs at 4K. We studied Co/Ag and Co/AgSn multilayers. The samples were made in a dc magnetron sputter deposition system with a computer controlled substrate positioning and masking apparatus. A SQUID based circuit with a high precision current comparator is used for the CPP MR measurements.

We find that CPP MR > CIP MR.

We also show how to analyze our CPP MR data. For Ag thicknesses 6nm and larger, and Co thicknesses 20nm and smaller, we find that a two channel series resistance model gives a good description of most of our CPP data. In this model, the spin up and spin down electrons are taken to carry current separately. A fit with no adjustable parameters showed that the resulting resistivities agreed with independently measured numbers to within mutual uncertainties.

Our results establish the CPP MR as an important complement to the CIP MR and show the importance of the F/N interface resistance to the giant MR.

To my parents

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## CHAPTER 1

## Introduction

A lot of interest has been focused on the unexpectedly large negative magnetoresistances (MR) -- magnetic field dependence of the resistance -- of artificial multilayers composed of at least one ferromagnetic component. The interest is because of both the underlying physics of this inhomogeneous system, and potential applications in industry, like position sensors or miniature reading heads for computer hard drives. In this dissertation, we describe our work on this topic, with emphasis on resistance when the measuring current is perpendicular to the multilayer planes.

In this chapter, we first review the historical development of multilayer studies, then briefly survey work on magnetic coupling and electron transport in *magnetic* multilayers.

#### **1.1 Overview of Multilayer Studies**

The concept of an artificial multilayer can be found in the literature as early as the 1920's. Koeppe (1923) (referred to in Deubner) and Deubner (1930) [1] made multilayers with the motivation to fabricate soft X-ray diffraction gratings. By sequential electrodeposition, Koeppe made Cd/Ag multilayers, but was not able to see diffraction patterns. Using the same method, Deubner observed diffraction patterns with Au/Ag and Ag/Cu multilayers, and suggested that physical vapor deposition might yield better structures.

More recently, there has been a rapid rise of interest in multilayers due to the advent of powerful new deposition techniques. The most commonly employed multilayer preparation techniques include: thermal evaporation; dc or rf sputter deposition; and molecular beam epitaxy (MBE); etc. All these techniques involve a vacuum chamber. The materials are heated in various ways to cause evaporation or, in the case of sputter deposition, the materials are bombarded by energetic ions (e.g. Ar) and sputtered onto substrates. Among these techniques, MBE systems are the best in terms of the ability to control and investigate sample quality *in-situ*. An MBE system involves an ultrahigh vacuum chamber, several effusion (or electron beam) cells, and an array of surface analytical instruments. An MBE-grown sample is usually deposited very slowly and the crystal growth is monitored all the time. Other sample preparation techniques usually produce samples with lesser crystalline quality at a faster pace.

To make multilayers with artificial periodicity, there must be material flux modulation control in the vacuum chamber, such as shutters to open and close the path between sources and substrates, and/or a mechanism to translate substrates between material fluxes. The different materials in a multilayer usually have different lattice structures and/or lattice constants. Single crystal multilayers, composed of different materials that grow epitaxially, usually need special recipes, such as certain ranges of substrate temperatures and selected buffer layers.

For the purposes of this dissertation the name *superlattice* is reserved for "multilayers with single-crystal structure". The name *multilayer* refers to a polycrystalline structure.

In the 1950's, multilayers containing ferromagnetic metals appeared with the motivation of achieving low a.c. losses or controlling coercivity [2-5]. It was already known in the 60's that magnetostatic coupling can direct the spins of neighboring ferromagnetic layers in opposite directions [4].

Superconducting properties of multilayers were investigated starting in the late 1960's. Much of the motivation was to test different mechanisms that might lead to structures with higher transition temperatures [6, 7].

Esaki and Tsu [8] proposed growing single-crystal multilayers of two different semiconductors in 1970. Significant modifications of the electronic properties were found as a result of the imposed periodicity [9]. The study of semiconductor superlattices is now an independent subject.

#### **1.2** Magnetic Multilayers

Magnetic multilayers, involving at least one ferromagnetic (F) component and non-magnetic (N) or antiferromagnetic materials, have been a topic of great interest since the late 1970's. A large literature has evolved on various ferromagnetic/nonmagnetic, ferromagnetic/antiferromagnetic, and helical magnetic/non-magnetic multilayers [9, 10]. Both theoretical and experimental efforts were made to study the surface and/or interface magnetism, interlayer magnetic coupling, and the effects of reduced dimensionality in magnetic multilayers. The possibility of exchange coupling of magnetic layers across non-magnetic layers had been considered for a long time before it was first clearly identified and characterized in magnetic rare earth superlattices, e.g. couplings of adjacent magnetic layers oscillate between ferromagnetic and antiferromagnetic depending upon nonmagnetic layer thicknesses in Gd/Y [11]; spiral magnetic coupling occurs in Dy/Y superlattices [12]; and antiferromagnetic coupling occurs in Fe/Cr structures [13, 14].

Together with magnetism studies, transport properties of magnetic multilayers have also been investigated. Unusual magnetoresistive effects were reported by this group at MSU and others [15, 16]. However, it was A. Fert and co-workers who discovered that there are very large negative magnetoresistances in MBE-produced Fe/Cr superlattices when an antiparallel coupling between Fe layers at zero field is present. The electrical resistivity was reduced to half of the original value when a high magnetic field was applied to a particular sample [17]. This effect was called giant magnetoresistance (MR). A similar effect was reported around the same time by Binash *et al.* for MBE-produced Fe/Cr sandwiches [18].

Subsequently, Parkin *et al.* not only reproduced Fert's results in sputter deposited polycrystalline samples, but also observed unexpected long range oscillations in the magnetoresistance as the thickness of the spacer layers was varied. The same oscillations were also observed in the magnetic exchange coupling [19, 20]. Since then, a wide variety of different magnetic multilayers have been studied. Later in this chapter we will briefly review these studies.

All the early studies on MRs were measured with current in the layer plane (called CIP geometry) and the magnetic field also usually in this plane. While the giant CIP MRs of different multilayer systems show diversity in details, two elements seem common:

- 1. The CIP MR depends on the relative alignment of the in-plane magnetizations  $\mathcal{M}_i$  of neighboring F layers. It is generally presumed to be due to a complex process involving transfer between F layers of spin-polarized electrons that encounter spin-dependent scattering, both within the F layers and at the F/N interfaces.
- 2. The largest CIP MR occurs when application of a magnetic field H directed in the layer planes causes the alignment of adjacent layers to change from antiparallel at low field  $H \approx 0$ , to parallel above the saturation field,  $H_s$ .

The CIP geometry contains inherent complications for quantitatively separating bulk from interface scattering, because the current density in the sample is nonuniform, and the current channels in the two different metals are connected by transmission of electrons through the interfaces. Such measurements led to disagreements over the question of the relative importance of bulk and interface scattering in giant MR [21, 22]. Recent developments on this issue are described in later section (see page 12).

We will argue in chapter 4 that the preferred geometry for separating bulk from interface scattering is one where the current flows perpendicular to the layer planes, the CPP geometry, since the current density is uniform across the sample area, and the electrons pass sequentially through the individual metals and the interfaces. We have fabricated samples on which both the CPP and CIP MR can be measured. The Co/Ag system was chosen to be investigated first because (1) these metals are mutually insoluble [23], thus we expect sharp, stable, and well defined interfaces; (2) previous work on this system in this laboratory [24].

In chapter 2 we describe how we produce and characterize our samples. All the samples for this dissertation were made by sputter deposition and characterized by X-ray diffraction.

The CIP geometry is easy to measure. The sample width and length can always be designed to make the resistance big enough so that a standard voltage measuring technique can be used. To measure resistance in CPP geometry, there are two feasible ways. With modern lithography techniques, one can shape the thin film sample to a very small width comparable to its thickness so that the resistance is reasonably big (say,  $10^{-2}\Omega$ ) [25]. The other way is to measure the small resistance (~  $10^{-7}\Omega$ ) of a film resulting from small thickness and relatively big area. We chose the latter way because we have a unique setup, which combines a Superconducting QUantum Interference Device (SQUID) and a high precision current comparator, to measure small resistances with high precision [24, 26]. This method is limited to the operating temperature of the SQUID. Our setup is described in chapter 3. Chapter 4 is devoted to the theory of magnetic coupling and of magnetoresistance.

Our results are presented in chapter 5, where we compare our CIP and CPP MR data, and analyze the CPP data in terms of a two channel series resistance model.

#### **1.2.1** Studies of Magnetic Coupling in Magnetic Systems

Since interlayer coupling was first found in rare earth and Fe/Cr systems, a wide range of spacer materials has been reported to mediate such coupling: (1) antiferromagnetic transition metals (Cr, Mn) [13, 14, 17, 19, 27–31], (2) non-magnetic transition metals (Ru, Mo, Pd, etc.) [19, 32–35], and (3) noble metals (Cu, Ag, Au) [20, 36–50].

For a quantitative evaluation of the exchange coupling per unit area between two ferromagnetic layers  $F_1$  and  $F_2$ , the exchange energy can be written in the form:

$$E_{1,2} = J_{1,2} \frac{\mathcal{M}_1 \cdot \mathcal{M}_2}{|\mathcal{M}_1| |\mathcal{M}_2|} = J_{1,2} \cos \theta_{1,2}$$
(1.1)

where  $\mathcal{M}_1$  and  $\mathcal{M}_2$  are the magnetizations,  $\theta_{1,2}$  the angle between them, and  $J_{1,2}$ the coupling constant per unit area.  $J_{1,2}$  depends on the properties of the spacer material, like thickness, crystal orientation, etc. With this definition,  $J_{1,2}$  is positive for ferromagnetic and negative for antiferromagnetic coupling.

Coupling that can be described by the form of equation (1.1) is called *bilinear* coupling. It has been observed experimentally that, under certain circumstances, coupling between two ferromagnetic layers favors a perpendicular alignment between  $\mathcal{M}_1$  and  $\mathcal{M}_2$ . This coupling is referred to as *biquadratic* coupling, which can be expressed as

$$E_{1,2} = B_{1,2} \cos^2 \theta_{1,2} \qquad \text{with} \quad B_{1,2} > 0$$

Theoretical models are proposed to calculate the coupling constants from either

total energy between the ferromagnetic layers, or from perturbative models. See chapter 4.

#### **Experimental Techniques**

Experimental techniques used to investigate the interlayer exchange coupling include: (1) magnetometry, magnetoresistance, and magneto-optics; (2) magnetic domain microscopy; and (3) other techniques. See reference [10] for a review.

(1) Magnetometry, magnetoresistance, and magneto-optics have been the most widely used methods to investigate interlayer coupling. One can measure antiferromagnetic coupling simply by applying external magnetic fields to align the magnetic moments. In the case of ferromagnetic coupling, Parkin and Mauri [32], and Fert *et al.* [51] used strong antiferromagnetic coupling to pin down one magnetic layer in order to measure the weaker ferromagnetic coupling. The Magneto-optical Kerr Effect (MOKE) provides the advantage that instead of probing the sample as a whole, it probes the sample locally. This feature allows one to investigate samples with wedge-shaped spacer layers, i.e. spacer layers with continuously increasing thicknesses. Unlike other methods, where one has to produce a large number of samples and determine spacer thicknesses for all the samples, using MOKE on wedge-shaped samples allows one to measure the coupling oscillations with high thickness resolution [29, 31, 47, 48].

(2) Magnetic domain imaging techniques such as scanning electron microscopy with polarization analysis (SEMPA) and Kerr microscopy have been used to image the magnetic domains. These methods do not measure directly the coupling strength but merely its sign. The first evidence of short-period coupling (about 2 mono-layers) was seen by applying SEMPA on wedge-shaped Fe/Cr/Fe sandwiches [28].

(3) Ferromagnetic resonance (FMR) and Brillouin light scattering (BLS) rely on measurements of spin-wave frequencies of the optical and acoustic modes [13, 36, 39, 40, 44]. One can obtain coupling strengths from these measurements [40, 53]. BLS measures a sample locally but FMR measures a sample as a whole. In the presence of antiferromagnetic coupling in magnetic multilayers, the magnetic unit cell is twice the chemical cell. Neutron scattering has proved the presence of such coupling [27, 34, 38].

#### **Experimental Results**

The Fe/Cr system was used as a "base system" to study magnetic coupling by different groups, since antiferromagnetic coupling was first reported by Grunberg *et al.* between Fe(001) layers separated by a Cr(001) spacer layer [13]. Other reports confirmed this observation [14, 17]. All these early reports seemed to show the coupling decreasing monotonically with increasing spacer thickness.

Parkin et al. then made a major step. They grew Co/Ru, Fe/Cr and Co/Cr multilayers by sputter deposition and observed coupling oscillations from antiferromagnetic to ferromagnetic with decreasing amplitude as the spacer layer thickness increased [19]. The oscillation periods are 12Å across Ru spacer layers and about 18-21Å across Cr layers.

In the early stage, models could not account for the long period oscillations. Instead most of the models found a rather short period of one or two monolayers which is no more than 5Å. A brief review of the theories and how this discrepancy can be resolved will be presented in chapter 4.

Unguris et al. [28], and Purcell et al. [29] then observed coupling oscillations with a period of about two monolayers superimposed upon the long range oscillation in epitaxial Fe/Cr/Fe(001) sandwiches where the Cr layer was wedge-shaped. It was found that good crystal quality was very essential to observing the short period oscillations. Disorder in the crystalline structure tends to smear out the short period oscillation and leaves only the long period. By imaging the domain structure of the same kind of sandwich, Ruhrig et al. reported a biquadratic coupling, i.e. the moments of the two Fe layers prefer to align perpendicular to each other, in the region between ferromagnetic and antiferromagnetic coupled parts of the sample [30]. The strength of the biquadratic coupling is at least one order of magnitude smaller than the bilinear one. Thus only when the bilinear coupling is close to zero, does the biquadratic one dominate.

Antiferromagnetic coupling and oscillatory coupling with increasing spacer layer thickness were also reported in other superlattice systems like Fe/Mn(001) [31], Fe/Mo(110) [35], Fe/Pd(001) [36], Fe/Cu(001) [41], Fe/Ag(001) [37], Fe/Au(001) [47], Co/Cu(001) [42, 48], and Co/Cu(111) [50]. A systematic study of interlayer coupling in sputter deposited Co-based multilayers was reported by Parkin [33]. For a review, see reference [54].

For the Co/Ag system, which we studied, Parkin did not observe any antiferromagnetic coupling or coupling oscillations in sputter deposited samples. The Co thickness was fixed at 15Å in his work. Araki *et al.* reported both these effects in MBE grown Co/Ag samples. They prepared two sets of samples, one (002) superlattice samples grown on MgO (001) substrates, and the other multilayers grown on glasses, both with very thin Co thickness, 6Å, and Ag thicknesses varying from 4Å to 35Å. Both sets of samples had saturation fields which oscillated with increasing Ag thickness and antiferromagnetic coupling was confirmed by Ferromagnetic Resonance. For Co thickness 15Å and larger, as we are concerned with in this work, no antiferromagnetic coupling nor coupling oscillation has been reported.

In the early stage, the coupling strength was shown to be constant experimentally, irrespective of the ferromagnetic layer thickness, see e.g. [55]. However, theoretical predictions by Bruno [56] and Barnas [57] showed that the coupling may oscillate with the ferromagnetic layer thickness. Such behavior was recently reported by Bloemen *et. al.* [58].

#### 1.2.2 Studies of CIP MR in Magnetic Systems

The giant MR was first found in exchange coupled systems [17, 18, 19, 59, 60]. Antiferromagnetic alignment of neighboring magnetic layers in these coupled sandwiches and multilayers can be changed to ferromagnetic alignment by applying a large enough external field.

Dupas et al. [61] and Chaiken et al. [62] subsequently showed that antiferromagnetic alignment could be obtained in systems with weak or no coupling. They fabricated multilayers with different coercivities for odd and even magnetic layers. By applying an external field to these samples to align all the magnetic moments, and then reversing the field to rotate those layers with low coercivity, one can create an antiferromagnetic alignment. With this design, the MR can be studied for systems that do not show antiferromagnetic coupling.

Dieny et al. [63] utilized strong pinning created by exchange anisotropy to constrain one of the layers in F/N/F sandwich samples. The other magnetic layer can be rotated by applying an external field. In this way they made possible magnetotransport studies of arbitrary ferromagnetic sandwiches exhibiting no interlayer coupling and also permitted a direct, quantitative measurement of the dependence of the resistivity on the relative orientation of the magnetizations. They called the change in MR with changing orientation of layer magnetization the "spin-valve effect".

The first theoretical model for giant MR was worked out by Camley and Barnas [64]. It is a semi-classical model starting from Boltzmann equation. The most complete quantum model is the one by Levy and co-workers. See chapter 4.

In the following, we briefly review experimental results on CIP MR.

#### **Dependence on Layer Thickness**

The dependence of the CIP MR on layer thicknesses in various multilayer systems has been studied extensively by different groups. Experiments on MR versus spacer layer thickness for Fe/Cr [19, 65] and Co/Cu [20, 66], for example, showed at least three oscillations with periods on the order of 10-20Å. The amplitude of the peaks decreases with increasing spacer thickness. At large spacer thickness, larger than 60Å for all multilayers reported, the oscillations die out and the MR decreases monotonically. The first and second peaks are strongly antiferromagnetically coupled, between second and third peaks there is a crossover to a weak coupling regime, and at large spacer thickness the layers are uncoupled.

The variation of the MR with the magnetic layer thickness was also studied. Dieny et al. [68] measured the MR dependence on the ferromagnetic layer thickness in their spin valve structures and compared the results with the theory by Camley and Barnas to determine the relative importance on bulk and interface scattering. Analysis of this kind was never completely conclusive on this issue. Okuno and Inomata [69] recently showed that CIP MR in Fe/Cr(100) multilayers oscillates with ferromagnetic layer thickness.

#### Variation with Field

The dependence of the CIP MR on the angle of the applied magnetic field was also studied [15, 17]. The longitudinal MR, which is measured with magnetic field along the current direction, and the transverse MR, which is measured with magnetic field in the layer plane but perpendicular to the current, has only slight difference. The difference between longitudinal MR and transverse MR is called the anisotropic MR (AMR). Chen *et al.* studied AMR in Co/Cu multilayers [70] and compared their results with a theoretical model [71]. When the magnetic field is applied out of the film plane, the MR behavior is more complicated [15]. The cause of this complex behavior is not well understood.

The magnetic easy axis for most magnetic multilayers that have been reported lies in the layer plane. However, when the magnetic layers are thin enough, the easy axis could be perpendicular to the layer plane due to anisotropy energy. Vavra *et al.* reported an easy axis perpendicular to the layer plane in their MBE grown Co/Au superlattice with Co 5Å thick. The CIP MR with field perpendicular to the layer planes, along longitudinal, and transverse directions are also reported [72].

#### Variation with Temperature

As the temperature increases, the MR ratio decreases [20, 43, 73, 74]. The effects contributing to the MR variations with temperature are: (1) there are additional inelastic scattering processes at higher temperature, like phonon and magnon scatterings etc., resulting in shorter mean free paths; (2) electron-magnon scattering introduces additional spin-flip scattering; (3) the spin asymmetry factors of the scattering processes are different between low temperature elastic processes and temperature induced scattering processes. Studies of temperature variations on different multilayer systems show a strong dependence in Fe-based systems [73] but only moderate in the Co/Cu system [20, 43]. Theoretical models have been worked out for the temperature dependence of MR [75, 76]. More comparison between experiments and theories is needed to understand the different temperature dependence of various multilayers.

#### **Interface** Structure

The CIP MR depends significantly on the multilayer physical structure. Data on the same multilayer system prepared by various ways, e.g. different deposition methods, buffer layers, deposition rates, substrate temperatures,  $\cdots$  etc., showed qualitatively similar, but quantitatively different, results. The reason is that resistivity is not an intrinsic property of a metal. Resistivity depends not only on intrinsic band structures but also on properties and concentrations of defects, impurities, and other scattering centers. Thus MR can be very sensitive to the sample growth conditions. Interface structure is also strongly dependent on the growth conditions. The influence of interface roughness on the MR in different multilayer systems has been studied by intentional mixing during growth or by annealing. Characterization of interface roughness was done by low angle and/or high angle x-ray diffraction. Reports from different labs on the correlation between roughness and MR were not conclusive [73, 77, 78].

To study the interfacial scattering processes in detail, selected elements have been inserted at the multilayer interfaces [21, 22, 79]. Depending on the multilayers and the inserted elements, MR effects increase in some cases and decrease in others. Spin dependent scattering was used to explain the experimental results [80]. These studies were also used to show qualitatively the relative importance of bulk and interfacial scatterings. The importance of interfacial scattering on MR in Fe/Cr was reported by several groups [73, 77, 78]. Dieny *et al.* argued that bulk scattering was the dominant mechanism for the spin-valve effect in permalloy and Ni sandwiches [63]. However, Parkin argued that, for Co or permalloy based sandwiches with a variety of spacer layers, scattering from magnetic states predominantly localized at interfaces is responsible for MR [22].

#### **1.2.3** Studies of CPP MR in Magnetic Systems

The first measurements of the CPP MR in magnetic multilayers were made by this group at MSU [81] and will be presented in this dissertation. We used superconducting materials as current and voltage leads to ensure uniform measuring current. Although the sample geometry is well defined in this technique, it is limited to low temperature because of the superconducting materials.

Gijs et al. [25, 82] subsequently measured the CPP MR on samples with small areas prepared by lithography. They used normal metals as electrical leads, and had to correct for the contribution from finite contact resistances. With this design they were able to measure the temperature variation of the CPP MR in Fe/Cr [25] and in Co/Cu [82] from 4K to 300K.

#### **1.2.4** Studies of Giant MR in Heterogeneous Systems

A giant MR effect in thin films containing magnetic granules, e.g. Cu films with Co precipitates, was reported simultaneously by Berkowitz *et al.* [83] and Xiao *et al.*[84]. The sizes of the magnetic granules ranged from a few to a few tens of nanometers. Both the MR and saturation fields depended on the granule size. These results demonstrated that the giant MR is not restricted to multilayer structures. Zhang [85] worked out a theoretical model for this heterogeneous systems and found that it is closer to the CPP MR of multilayer systems than to the CIP MR.

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## CHAPTER 2

# Sample Fabrication and Characterization

#### 2.1 Introduction

Our multilayers were made with a UHV compatible dc magnetron sputtering machine. With an *in-situ* mask changing system and a cooling system, we could control sputtering conditions and make complicated samples without breaking the vacuum. The sample shape was designed to permit simultaneous measurement of both CPP and CIP MRs (Figure 2.1). Some equivalent simple square multilayers with nominally the same bilayer thicknesses were also deposited for ease of magnetization measurements and for cross-checking the CIP MR.

All the samples presented in this dissertation were made under narrowly defined sputtering conditions. The Ar pressure was kept the same, and the sputtering gun powers for the individual materials were kept closely the same. The anticipation was that sample quality would be similar. The thicknesses of the multilayer constituents as well as the total thickness of the multilayer were varied systematically. Structure variations induced by different deposition conditions are not a subject of this work.

The MRs of the samples were measured first, in both the CIP and CPP geometries.

The measuring techniques will be described in the following chapter. X-ray diffraction studies were then made to check the bilayer thickness and to examine the structure of the sample. Finally the sample geometry was determined by a surface profiler to provide information necessary for quantitative analysis.

#### 2.2 Sample Fabrication

The sputtering system was described in detail in the dissertation of J. Slaughter [24]. A cryopump (CTI Cryo-Torr 8 [86]) provides high pumping speeds (1500 l/s air, 4000 l/s water vapor) with no oil contamination for the system chamber. Four L.M. Simard [87] "Tri-Mag" sputtering sources accommodate up to four targets. On top of the targets, four separate sets of shutters can be closed or opened at the same time to interrupt the particle streams or allow them to go through. A rotary plate (called the substrate positioning and masking apparatus, SPAMA) holds the substrates and the masking system above the shutters. A stepping motor, controlled by computer, positions the substrates above the desired sources.

Two major modifications were made since Slaughter's dissertation was written:

- A cooling system designed and constructed by C. Fierz and W.P. Pratt Jr. was added [88]. The whole SPAMA plate is cooled by high pressure Nitrogen gas passing through a capillary, which goes through a heat exchanger connected to a Meissner trap (liquid Nitrogen trap) in the sputtering chamber and then through the shaft that holds the rotary SPAMA plate. The high pressure Nitrogen gas, from a commercial gas cylinder, is controlled by a regulator at the pressure 800-1,100 psi.
- 2. A newly designed *in-situ* mask changing system replaced the old one. Substrates are now mounted in holders that fit into circular holes in the rotary plate. Each


Figure 2.1. Sample design for CPP and CIP MR measurements.

substrate has an 2-inch diameter mask attached beneath the substrate holder. These masks accommodate three different shapes: a 1.1mm wide strip; another 1.1mm wide strip perpendicular to the first one; and a sample shape consisting of a 4mm diameter circle with two 0.3mm wide strips on the sides and rectangles at the ends, see Figure 2.1. These masks can be rotated in and out of place under the substrates *in-situ* by a wobble stick that extends outside the vacuum chamber.

Our Nb and Co targets were bought from commercial companies [89]. Initial Ag targets were also bought, but we now make them using an rf-furnace. We melt Ag powder in a cylindrical graphite bowl painted with boron nitride. The bowl is placed on a stand inside a quartz tube which slides down through the rf coils. The tube is pumped down with a diffusion pump to the  $10^{-6}$  torr range, then backfilled with 1/3 atm of Ar mixed with 2% H<sub>2</sub> just before the rf coils are turned on. Upon cooling, we obtain a slightly oversized Ag disc, which is then cut to final shape with a traveling wire EDM (electrical discharge machine). A AgSn target was made the same way with 4 at.% Sn added to the Ag. This target was used to make samples with AgSn spacer layers between Co layers to test for mean free path effects.

A typical sputtering run starts with cleaning all the components to prevent

contamination. The metal deposits from previous runs are removed by razor blades and appropriate chemicals (acids or bases) [24]. Just before assembly, the components and the substrates are cleaned with acetone and ethanol. The system is then assembled, closed, and pumped down with a gentle bake at about 80°C. If the sputtering run is the first of a series, the bake lasts for 10 hours. If the system is only opened for a short time to change substrates, the bake lasts only 4 hours.

A base pressure of  $1-2 \times 10^{-8}$  torr or lower is always obtained with the Meissner trap cooled before sputtering. During sputtering, the pressure of high purity Ar gas in the chamber is  $2.5 \pm 0.3$  mtorr. The substrates are  $12 \pm 1$  cm above targets. The voltage and current of the sputtering gun for each material are kept fixed. Deposition rates are measured by quartz crystal film thickness monitors. From past comparison with x-ray results for bilayer thicknesses, and with a surface profiler for total thicknesses, the actual deposition rates are 4% higher than the monitor reading. The thickness monitor reading plus four percent is then used to control the exposure time of substrates to the desired material flux. In each sputtering run, rates stabilize after about five minutes' warm-up. For different target materials bought from different companies over the years, and for different target thicknesses, fixed voltages and currents have resulted in ranges of deposition rates: for Ag  $\sim 11-14$ Å/sec; Co  $\sim 8-$ 10Å/sec; and Nb  $\sim 9-10$ Å/sec. With the help of the cooling system, the substrate temperatures are kept between -30°C and 30°C while the multilayers are deposited. If the temperature becomes too high, the guns are shut off to allow the system to cool.

The samples on which the CIP MR and CPP MR are measured simultaneously are deposited on c-axis oriented,  $1.27 \text{ cm} \times 1.27 \text{ cm}$  sapphire substrates. We first deposit a Nb strip on the substrate, followed by the multilayer, and then another Nb strip perpendicular to the first one. Two different procedures were used to make these samples. The major difference was the amount of time the bottom Nb strip surfaces was exposed to ambient pressure in the chamber before the multilayers were deposited. Procedures, A and B, were:

Procedure A:

- 1. With the mask closed, measure the deposition rate of Nb.
- 2. Open the mask to bottom strip position, deposit 5000Å of Nb.
- 3. Close the mask. Measure the deposition rates of Co and Ag.
- 4. Open the mask to multilayer position, deposit the multilayer.
- 5. Change the mask to top strip position, deposit another 5000Å of Nb.
- 6. Close the mask.

Since the deposition rates were measured between the bottom Nb strip and the multilayer, the exposure time of the bottom Nb to the ambient pressure was about 5-6 minutes. The top Nb strip was deposited immediately after the multilayer, with an exposure time less than 2 minutes.

Procedure B:

- 1. With the mask closed, measure the deposition rates of all materials.
- 2. Open the mask to bottom strip position, deposit 5000Å Nb.
- 3. Change the mask to multilayer position, deposit the multilayer.
- 4. Change the mask to top strip position, deposit another 5000Å Nb.
- 5. Close the mask.

In this procedure, the exposure times of both the bottom Nb and the multilayer surfaces were less than 2 minutes. Most of the data presented here were taken on samples made with procedure A. We made samples with procedure B to double check our data. It will be shown in section 3.2.3 that the extra time in procedure A did not change the results significantly.

Our new masking design worked well at first, but repeated cleaning of the aluminum sample holders with acid gradually etches the holders. The initial half-inch square holes fitted the substrates nicely. But if the holes become bigger, when a multilayer is being deposited, the motion of the whole rotary plate can cause the substrate to shift. Thus, some of our samples have multilayers slightly displaced into two parts. The CPP samples are not affected because the CPP samples are at the center of the multilayers where the layering is still good. On the other hand, since the CIP samples are only 0.3mm wide, excessive displacement can ruin a sample. Such samples will be marked "bad" and footnoted as "misaligned sample" when we present our data for CIP MR.

For some of our multilayer samples, simple rectangular multilayers with nominally the same bilayer thicknesses were also deposited, a few onto sapphire substrates and most onto (100) Si substrates because Si is cheaper and easier to break into halves. These films were used both to test for substrate dependence and to measure the sample magnetizations. Most of these samples were deposited directly onto the substrates. We deposited a few samples onto Nb buffer layers in order to test the effect on both the CIP MR and the magnetization.

## 2.3 Sample Structures

#### 2.3.1 X-ray Diffraction

X-ray diffraction was used to study multilayer structures. From Bragg's law,

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 $n\lambda = 2d\sin\theta,$ 

we know that an incident x-ray with wave length  $\lambda$ , gives constructive interference for a periodic distance d in the sample at equal incident and detection angles  $\theta$ . These angles are measured from the sample planes. In our multilayers, there are three separate lengths: the Co lattice constant, the Ag lattice constant, and the bilayer thickness  $\Lambda$ . Because the lattice constants for bulk Co and Ag are the same order of magnitude as  $\lambda$ , constructive interferences from bulk Co and Ag show up as high angle peaks. Constructive interference associated with the additional bilayer thickness gives both low angle peaks and satellites around the bulk high angle peaks. A more extensive study of x-ray diffraction of Co/Ag multilayers, including a theoretical step model and experimental analysis, was given in the dissertation of J. Slaughter [24].

From Slaughter's work we learn the following, which was confirmed by our studies. Our sputter deposited multilayers have polycrystalline structures. The dominant growth orientation of each material is the one with the most dense crystallographic planes parallel to the substrate surface. The Nb films have bcc structure with the (110) planes parallel to the substrate, Ag has fcc (111), and Co has either fcc (111) or hcp (0001). A step model simulation for samples with equal Co and Ag thicknesses has been work out in three different cases: hcp-Co (0001), fcc-Co(111), and bcc-Co(110) layered with fcc-Ag (111). When the bilayer thickness is less than 30Å, the three cases are similar. For bilayer thicknesses between 30Å and 80Å, the fcc-Co simulations agree with the form of experimental spectra slightly better than the hcp-Co ones. For bilayer thicknesses larger than 80Å, hcp-Co simulations are slightly better. The rocking curves of our samples have full widths at half maximum of ~ 12°. This shows that there is a distribution of crystalline orientations in our samples, the majority of which lie within  $\pm 6^\circ$  of the preferred direction.

All of the x-ray diffraction measurements were made on a Rigaku diffractometer [90] with a Cu-K<sub> $\alpha$ </sub> rotating anode ( $\lambda = 1.5418$ Å). Low angle diffraction peaks were seldom seen, probably because of curved sapphire substrates and columnar growth.

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Figure 2.2 shows a typical high angle diffraction scan on a simple square  $Co(60\text{\AA})/Ag(60\text{\AA})$  sample (bilayer thickness  $\Lambda = 120\text{\AA}$ ). The peaks at detector angle 2 $\theta$  roughly 38 and 44 degrees are constructive interferences from bulk Ag and bulk Co respectively. The bulk Nb peak is also around 38 degrees. These angles verify the crystal orientations given above. The satellites around the bulk peaks are constructive interferences from the multilayer periodicity. Knowing these satellites' positions, we can calculate the bilayer thickness from the Bragg condition. In Figure 2.3, the slope of the best fit line is the best value for  $\Lambda$ . All the bilayer thicknesses derived from x-ray diffraction were within 4% of the desired thicknesses, thus confirming our sample fabrication procedures. X-ray results of all the samples presented in this dissertation are listed in appendix B.

The dashed line in Figure 2.2 is a fit with SUPREX, which is a more sophisticated computer program by Fullerton et al [91]. A project using this program to fit x-ray spectra of different multilayers is currently being carried out by L. Su at MSU. There are several models in the program, like superlattice, quasicrystal with strain, high  $T_c$  superlattice,  $\cdots$  etc. Within each model, there are at least 14 fitting variables to take into account interface roughness, lattice spacing variations, interdiffusion, etc. The fit shown in the figure is by a superlattice model for fcc Ag and hcp Co, without any interdiffusion. The spacing between the Co and Ag layers was taken to be the average of the two lattice spacings, with strain allowed for three monolayers on either side of the interface. Fluctuations of individual layer thicknesses were allowed to be 1.25 monolayer for Ag and 1.5 monolayer for Co. The output file, including all input and output parameters, for this fit are given in appendix B. We see that most of the features of the experimental spectrum are nicely reproduced. The fit supports independent measurements suggesting that our "interface diffusion zone" is usually no more than 2-4 monolayers thick (see, e.g. [92, 93]). The bilayer thickness from the fit is  $114.8 \pm 6.3$ Å, consistent with the number derived as shown in Figure 2.3. More



Figure 2.2.  $\theta$ -2 $\theta$  spectrum for  $\Lambda = 120$ Å,  $t_{Co} = t_{Ag} = 60$ Å. The dashed line is a fit, see text.



Figure 2.3. Determination of bilayer thickness  $\Lambda$  from Bragg's Law. The slope of the straight line is the bilayer thickness.

such studies are currently underway.

#### 2.3.2 Other Studies

Other structural studies made on Co/Ag multilayers produced by our sputtering system include Cross-section Transmission Electron Microscopy and Nuclear Magnetic Resonance. We briefly summarize what has been learned in this section. Some of these multilayers were grown under different deposition rates from our samples, but the general conclusions should be applicable.

#### **Cross-section Transmission Electron Microscope Studies**

Figure 2.4 shows a schematic drawing of a Cross-section Transmission Electron Microscope image. This project was carried out by D. Howell *et al.* [94]. The multilayer sample was taken off the substrate and sliced into thin slices for imaging. Preliminary results show that the first few bilayers have uniform layering. When the samples get thicker, columnar growth begins, leading to a more complex structure. Within each column the layering is rather uniform, but the layering between different columns does not always register.

#### **Nuclear Magnetic Resonance Studies**

Nuclear Magnetic Resonance (NMR) studies on Co/Ag multilayers produced in our sputtering system were performed by van Alphen *et al.* [95]. In a NMR spectrum, bulk hcp Co has a resonance frequency close to, but slightly higher than that for bulk fcc Co. Thus, from the line shape of the main peak, one can compare qualitatively the relative amount of hcp versus fcc Co in different samples. It has been found that when a Co atom has its nearest neighbors replaced by other atoms, the resonance frequency decreases, giving rise to satellites in the NMR spectrum. The satellites of a multilayer are attributed to Co atoms at interfaces. From the relative intensities



Figure 2.4. Schematic drawing of a Cross-section Transmission Electron Microscope image.

between the main peak and the satellites, one can infer the growth mode (island versus layer by layer) and interface roughness. One can also derive the hyperfine field from the resonant frequency and resonant field. The relative change in hyperfine field can be related to the relative change in atomic volume. One can thus infer the strain in the multilayer sample.

NMR studies on Co/Ag multilayers produced at MSU show the following. Both the main peak and the satellites are rather broad compared with Co/Cu multilayers prepared by e-beam evaporation in UHV. Thus the Co layers in Co/Ag multilayers are mixtures of fcc Co, hcp Co, and stacking faults. From the analysis of the intensity ratio between the main peak and the satellites, it is inferred that Co layers are three dimensional islands for nominal thicknesses of 10Å or less, but become continuous for thicknesses larger than 10Å. Assuming the relative amounts of fcc Co, hcp Co, and stacking faults are the same in all samples, one can attribute the shift in resonance frequency to the strain in the samples. Analysis of the hyperfine field versus different thicknesses shows that the strain is proportional to  $1/t_{Co}$  and is independent of  $t_{Ag}$ . To conclude, we have some preliminary results on our multilayer structures. However, it is not easy to extract quantitative information useful for resistance measurements, especially about interfaces. More structure analysis is underway. We will show in chapter 3 that our resistance measurements are reasonably reproducible.

For future studies of our multilayer structures, the interfacial roughness and the degree of roughness correlation can be determined from combined x-ray measurements of (1)  $(\theta, 2\theta)$  scan, (2) rocking curve scan, and (3) offset  $(\theta, 2\theta)$  scan [96]. These various geometries map out the scattered intensity distribution in different cuts through reciprocal space. In the  $(\theta, 2\theta)$  scan, the specular reflection is measured on top of the diffuse background; the true specular reflection can be separated out in a rocking curve scan. In addition, the rocking curve profiles the angular distribution of the diffusely scattered x-rays perpendicular to the surface normal. In an offset  $(\theta, 2\theta)$  scan, one measures the x-ray spectrum with the sample misaligned by a small angle to avoid the strong Bragg peak. This measurement profiles the diffusely scattered x-ray intensity distribution along the surface normal. Combining the true specular reflection and the diffusely scattered intensity distribution, one can infer the interfacial roughness size and any correlations.

Several other techniques can also be employed for structure analysis. Transmission Electron Microscopy has been used to study columnar growth, the degree of crystallinity, etc. [94]. High Resolution Electron Microscopy (HREM) can be used to study interfacial structure [97]. It is important to combine x-ray diffraction and HREM. X-ray diffraction measures scattered intensities, so the phase information is lost. It also averages over the sample area. HREM provides local information. But HREM images result from phase contrast, thus it is impractical to determine sample structure uniquely. It also averages over the path length of the electron through the sample. However, by combining quantitative intensity measurements with structural modeling, it is possible to perform a detailed structure characterization. X-ray absorption spectroscopy (XAS), and in particular extended X-ray absorption fine structure (EXAFS), can be used to probe the local structure of the multilayer constituents [98]. Recent development of the magnetic force microscope (MFM) also allows imaging magnetic structures with a resolution about 10nm [99]. Such studies would improve our understanding of both the physical and magnetic structures of our multilayers.

# 2.4 Sample Geometry

To check the thicknesses of our sputter deposited samples and the widths of the Nb cross-strips, we used a Dektak IIA surface profiler [100]. This instrument measures the profile of the sample by passing the sample beneath a diamond-tipped stylus. Surface variations cause the stylus to be translated vertically. Stylus movement is converted into a digital signal and stored in a microprocessor for further analysis. There are built-in functions to determine distances between points, step heights, surface roughness etc. Results can be printed out. The vertical resolution is 0.5 nm and the horizontal resolution is 50 nm.

The largest source of error in determining the sample geometry is that associated with substrate imperfections such as curvature and slope discontinuities. Most of the sapphire substrates we used do not have slope discontinuities, but are slightly concave upward due to polishing. Since the film edges are not perfectly sharp, due to the small gap between the mask and the substrate while sputtering, it is not easy to determine the film thickness and Nb strip widths.

#### 2.4.1 Multilayer Thickness

The thicknesses of the films must be measured over some horizontal displacement. If the substrate was flat, we know the substrate position underneath the film. A curved substrate (see Figure 2.5) in the transition region makes it difficult to decide the substrate position, resulting in difficulty in determining the sample thickness and leads to an uncertainty of approximately  $\pm 3\%$ .

Another way to determine the multilayer thickness is from X-ray diffraction. The total thickness is just the average bilayer thickness, calculated from the spacing of the satellite, times the number of periods in the sample.

Both methods give thicknesses within 6% of the desired nominal thicknesses which were chosen when the samples were deposited.

#### 2.4.2 CPP Area

The effective CPP sample is the overlap area of the crossed Nb strips. To determine this area, the widths of both Nb strips need to be measured. The width of a strip was determined by finding the points on both sides of the strip where the Nb is  $\approx 10$ nm thick and measuring the distance between them. For samples on concave upward substrates, the lowest points were taken to be the bottom of the Nb strips. For some of the samples the substrates were concave downward, the vicinities where the slopes began to change significantly were taken as Nb bottom.

The 10nm criterion was chosen according to the literature [101] to take into account the effects:

- 1. A thin film has a suppressed  $T_c$ .
- 2. The superconductivity is destroyed by proximity effect due to the neighboring ferromagnetic Co.

This criterion was supported by later studies on Co/Nb/Co sandwiches [102].

Each Nb strip width was normally measured three times and the average was taken as the width. The difference between the average value and the farthest measurement



Figure 2.5. Dektak profile of a Nb strip. Notice the vertical scale is in Å and horizontal one is in  $\mu$ m.

was taken as the estimated uncertainty. The uncertainty for each strip was 1-3%, resulting in uncertainties for effective CPP sample areas of about 2-5%.

## 2.4.3 CIP Geometry

We will analyze our CIP data only qualitatively, in terms of the relative change in resistance  $\Delta R/R$ . No efforts were made to determine the CIP sample geometries.

# CHAPTER 3

# Measurement Techniques, Brief Results and Uncertainties

Our experimental setups for simultaneously measuring the CPP and CIP MRs, and for measuring magnetizations, will be described in this chapter. Some typical results will be shown, followed by discussion of the reproducibility of our data. Detailed data analysis will be given in chapter 5.

All the measurements presented in this dissertation are made at or near 4.2 K.

# **3.1** Experimental Setup

#### 3.1.1 CPP Geometry

Our CPP MR measuring system combines a high precision current comparator, capable of resolving current changes of parts in  $10^8$ , and a superconducting quantum interference device (SQUID) able to detect  $10^{-15}$  volts, to measure very small resistances with very high precision [26]. A schematic drawing of the circuit is shown in Figure 3.1. Since a medium precision is satisfactory for this work, a small inductor roughly  $50\mu$ H and a small resistor roughly  $170\mu\Omega$  were added in series to the SQUID input as a high frequency noise filter. The SQUID is a null detector in the circuit,



Figure 3.1. Experiment setup for CPP geometry resistance measurements.

where all the connecting wires are superconducting. The reference resistor has the value  $1.84 \pm 0.01 \mu \Omega$  [88]. The current comparator puts out two currents,  $I_1$  and  $I_2$ , and the ratio is adjusted to balance the circuit. When the circuit is balanced, the voltage drops across the reference resistor and the sample are the same. The ratio of the resistances is thus the inverse of the current ratio at balance. The same balancing procedure is repeated at each desired magnetic field.

Our CPP sample (Figure 2.1) consists of a bottom Nb strip, the multilayer of interest, and a top Nb strip. The effective CPP MR sample is the area where the two Nb strips overlap. At the measuring temperature, the Nb strips superconduct. The measuring current goes through one of the Nb strips, flows perpendicularly through the multilayer, and comes out the other Nb strip. The two constant potential Nb surfaces ensure that the current density is uniform, with the short and wide geometry of the sample ensuring that the fringing currents are small. For a more extensive discussion about current distributions in our CPP geometry, see appendix A. If we simply approximate the sample as a parallel capacitor, the lateral extent of the fringing current is of the order of the multilayer thickness. With a thickness  $1\mu$ m and an effective cross-section 1.1mm×1.1mm, the "fringing cross-section" is 1.102mm×1.102mm. Thus the experimental error due to the fringing currents is less than 0.4%.

The measuring current through the sample,  $I_2$  in Figure 3.1, is usually 50mA. During the years, we established a procedure to check whether the sample resistance was current-dependent at zero magnetic field. For each sample, we used selected sample currents among  $I_2 = 0.5, 5, 10, 25$ , and 50 mA etc. to check that the readings agree within uncertainties.

The solder joints to the Nb strips, at positions I and V in Figure 2.1, are made in two steps. First a thin layer of In is put on by ultrasonic soldering to break through the oxide on the Nb surface. Then the superconducting wire is soldered on with cerralloy 117 which has a low melting point 117°F and a superconducting transition temperature around 7K. The bond between In and Nb is strong, yet In is not superconducting at 4.2K (the  $T_c$  of In is 3.4K). This finite resistance of In does not affect the measurement because (1) the current comparator is a constant current source, so the In resistance does not change the measuring current; (2) there is no current flowing in the voltage circuit at balance, so the In resistance does not produce any voltage drop. The In resistance is negligible compared with the resistor we added in as part of the noise filter.

The total voltage measured across the sample is between the two superconducting Nb strips, and thus includes both the voltage drop across the multilayer and the ones across the two superconductor/non-superconductor interfaces. These extra interface resistances will be discussed later in this chapter (see Page 36).

#### **3.1.2 CIP Geometry**

Four-terminal measurements are used for the CIP geometry. The CIP resistance is dominated by the two thin strips between the center circle and the two pads of the multilayer. With a pair of current-voltage Cu wires on one of the sample pads and another pair of wires on the other pad (positions i and v in Figure 2.1), we use either a multimeter Fluke 8502A to measure the resistance or a SHE model PCB potentiometric conductance bridge to measure the conductance of the sample.

The advantage of the multimeter is that it can measure a large range of resistances. The last digit is  $0.01m\Omega$ , and the noise range is about  $0.1m\Omega$ . This is satisfactory for all the CIP samples in this work. The disadvantage of the multimeter is that it puts out 10mA measuring current continuously, and the power produced by this current can cause Joule heating. The constant voltage excitation of the conductance bridge can reduce this effect. The conductance bridge produces a 27.5 Hz,  $300\mu$ V square wave across the sample voltage terminals. The disadvantage of the conductance bridge is the smaller range it covers. The smallest resistance it can measure is  $0.05 \Omega$  and for resistances above 10  $\Omega$ , the reading is so small that the noise is about 10% of the reading.

The multimeter and the conductance bridge were cross-checked against each other in the overlap region to make sure that they agree to within mutual uncertainties. The conductance bridge is the first choice for CIP data so long as the reading doesn't go over range and the uncertainty isn't significant compared with the reading.

#### 3.1.3 Magnet

Our magnet is a hand wound small superconducting magnet capable of producing about 1 kOe (kilo-oersted) with 20 amperes. Using mono-filament NbTi superconducting wire with Cu cladding, we wound 40 turns of wire into two layers around two parallel  $\frac{3}{16}$  inch brass rods 2.5cm apart. These rods are threaded at one end to attach to the cryostat built by V.O. Heinen and W.P. Pratt Jr. The magnet wires were then potted with epoxy to prevent any wire movement. The resulting magnet is an oval-like cross-section, 1cm tall coil. The magnet power supply is a locally made low noise dc power supply operating on two 6 volt automobile batteries. To reduce field fluctuation due to the high frequency noise produced by the power supply, we added a small piece of thick Cu wire as a resistor between the magnet leads to increase the circuit time constant. Two layers of pure lead sheet ( $T_c = 7.2$ K) are wrapped outside the magnet's supporting rods as superconducting magnetic shields to protect the SQUID from magnetic fields. The cryostat fits in a four-layer glass dewar which holds liquid Helium and liquid nitrogen. A  $\mu$ -metal shield is wrapped around the outside of the glass dewar and all the apparatuses are housed in a screened room to reduce electro-magnetic interference.

The magnetic field was calibrated at the center of the magnet using a Hall probe which was in turn calibrated against a Quantum Design MPMS (see next section). A polynomial fit gives:

$$H = 0.45 + 43.4 I - 0.258 I^2 + 0.00692 I^3$$

where the magnetic field H is in Oersted and magnet current I in Amperes. A brass sample holder is placed inside the magnet and fixed by epoxy so that the CPP part of the sample is as close to the center of the magnet as possible. With this magnet we can only apply magnetic fields parallel to multilayer planes. The current in the CPP geometry is always perpendicular to the applied magnetic field and we also keep the current in the CIP geometry perpendicular to the applied field. The CIP part of the sample is then inside the magnet but off center. Since the magnet is only calibrated at the center, and the field inside the magnet is non-uniform due to its shape, we expect a larger systematic uncertainty in the field for CIP MR. The CIP MR measured on equivalent film in the Quantum Design MPMS will be used for comparison when we discuss issues related to the magnetic fields.

#### 3.1.4 Quantum Design MPMS

We use a Quantum Design Magnetic Property Measurement System (MPMS) to measure magnetization and CIP MR. The sample chamber of the MPMS is not big enough to accommodate our 1.27cm square sapphire substrates. Since sapphire is not easy to break, we deposit square multilayer comparison films onto 1.27cm square Si substrates (see section 2.2). The sample is then scribed with a diamond scribe from the back of the substrate and broken into two halves, one for the CIP MR and the other for magnetization.

The MPMS covers the temperature range from 2 to 400K. Liquid helium temperature, 4.2K, is an inconvenient working point because the system must be in its low temperature mode, where the 1K pot has to be refilled every 90 minutes or so and it takes 15 minutes for the temperature to stabilize again. To avoid this problem, we chose 5K as our measuring temperature.

Magnetization measurement is a built-in function in the MPMS. The sample is placed in a straw which is tied at one end for the sample to rest on. The straw is attached to the sample rod, and a program is written to control all the parameters. These measurements are made with the magnetic field parallel to the sample plane. Before measurements, the sample needs to be positioned at the center of the magnet. There is a function in the MPMS to ensure that the sample measurement path is symmetric with respect to the pick-up coil that couples to the SQUID. A small magnetic moment is required for this process but our sputtered multilayers do not all have spontaneous magnetizations, so we have to apply a small field, 20 Gauss, to center the sample. Thus for all the initial zero field magnetization data, the samples were exposed to a small magnetic field prior to the measurement.

To measure the CIP MR, we use a Keithley 224 current source and a Keithley 181 nanovoltmeter for four-terminal resistance measurements. The measuring current is 1mA and the magnetic field is in the sample plane and parallel to the current. The MPMS controls the temperature and the magnetic field. The current source and the voltmeter are controlled by a computer through an IEEE 488 interface, so we simply modify the computer program to communicate with the current source and voltmeter.

## **3.2** Experiment Results

Assuming a homogeneous sample and a uniform current distribution, the resistivity for a bulk sample is related to its resistance by,

$$R = \rho \frac{l}{A}$$

where l is the sample length and A is its cross section area. The resistivity is the quantity to use for comparing different bulk materials.

For boundaries or interfaces, which are two dimensional, the equivalent quantity is derived as follows. If the *conductance per unit area* of a boundary between two metals is designated as 1/r, then the total conductance,  $1/R_b$ , of an area A, is:

$$\frac{1}{R_b} = \sum_i \delta A_i \left(\frac{1}{r}\right) = \frac{A}{r}$$

The 'specific resistance' of the boundary is thus  $r = AR_b$ , the sample area times the boundary resistance.

The superconductor/non-superconductor interface properties are important to our CPP geometry measurement. Studies have been made in different labs about superconductor/normal metal interfaces [94-97]. Studies of superconductor/ferromagnet interfaces were made in this laboratory previously [107]. A series of S/F/S sandwiches composed of different thicknesses of F films between 500nm S strips were studied. Since the S layers have zero resistance, the area times CPP resistance is:

$$AR = 2AR_{S/F} + \rho_F t_F$$

where  $\rho_F$  is the bulk residual resistivity of the F metal and  $t_F$  is the F layer thickness. For Nb/Co/Nb sandwiches, it was shown that [107]:

- The Nb/Co interface has finite resistance. The CPP resistance times area versus Co thickness plot showed linear behavior and the extrapolation to zero Co thickness gave an intercept  $6.1 \pm 0.3 \, f\Omega m^2$ . This non-zero resistance,  $2AR_{Nb/Co}$ , is from the combined Nb/Co and Co/Nb interfaces. Similarly sputter deposited Nb films in these studies have resistivity around  $60n\Omega m$  just above transition temperature.
- The Nb/Ag/Co interface resistance is indistinguishable from the Nb/Co interface resistance. Ag thin films next to superconducting Nb have negligible resistances due to proximity effect. Up to 60nm thick Ag films were observed to have negligible resistances next to 500nm superconducting Nb films.
- The Co resistivity in the CPP geometry is comparable to its value in the CIP geometry. The slope of the AR versus t<sub>F</sub> plot, (52±3)nΩm, agrees with independent parallel resistivity measurements on sputter deposited Co single films, (58±6)nΩm. These large resistivities compared with those of the high purity metals are due to point defects and grain boundaries in the sputter deposited samples.

We independently measured the resistivities of several of our sputter deposited Co, Ag, AgSn, and Nb films just above its transition temperature with either Van der Pauw measurements [108] or the usual way of measuring resistance, thickness, width, and length of the film. The results are:

$$\rho_{Co} = 68 \pm 10 \,\mathrm{n}\Omega\mathrm{m}$$

$$\rho_{Ag} = 10 \pm 2 \,\mathrm{n}\Omega\mathrm{m}$$

$$\rho_{AgSn} = 221 \pm 36 \,\mathrm{n}\Omega\mathrm{m}$$

$$\rho_{Nb}(@12\mathrm{K}) = 72 \pm 9 \,\mathrm{n}\Omega\mathrm{m}$$

The slightly bigger value of  $\rho_{Co}$  compared with previous result is most likely the results of a different vendor of the Co targets [89].

The electron elastic mean free paths in our sputtered deposited Co and Ag films can be estimated as follows.

For Ag and AgSn films we use the free electron model, which gives resistivity  $\rho$  for a metal [109]:

$$\rho = \frac{m}{ne^2\tau}$$

where *m* is the electron rest mass, *n* the number of electrons per unit volume, *e* the electron charge, and  $\tau$  the electron scattering time. Since the conducting electrons lie close to the Fermi surface, we can write  $\tau = l/v_F$ , where *l* is the electron elastic mean free path and  $v_F$  the Fermi velocity. Thus we have:

$$\rho l = rac{mv_F}{ne^2}$$

Putting in for Ag  $v_F = 1.39 \times 10^6 \text{m/sec}$ ,  $n = 5.86 \times 10^{28}/\text{m}^3$  [109], and our measured resistivities, we find:

$$l_{Ag} = 84 \pm 17 \,\mathrm{nm}$$
  
 $l_{AgSn} = 3.8 \pm 0.6 \,\mathrm{nm}$ 

Co has a much more complex Fermi surface, where the free electron model is not a good approximation. From references [107, 110], the range of values for  $\rho l$  in Co are estimated as:

$$(\rho l)_{Co} = 0.7 \sim 2.3 \, \mathrm{f} \Omega \mathrm{m}^2$$

With our measured resistivity for Co, we get bounds for the mean free path:

$$l_{Co} = 9 \sim 40 \,\mathrm{nm}$$

#### **3.2.1** Resistance and Magnetic Moment

Figure 3.2 shows the CPP resistance, the CIP resistance, and the magnetic moment  $\mathcal{M}$  versus magnetic field H for a  $[Co(6nm)/Ag(6nm)] \times 60$  multilayer, i.e. Co layer thickness 6nm, Ag layer thickness 6nm, and 60 bilayers. The CIP resistance is measured simultaneously with CPP and the magnetic moment is measured on a separate, nominally identical film. The error bar of each data point in the figures is smaller than the symbol size.

Most of our multilayer samples show qualitatively the same behavior. At  $H_0$ , in the as-deposited sample before any field is applied (see Figure 3.2(b)), the multilayer has its largest resistance. The resistance decreases when a magnetic field is applied and reaches a minimum value at  $H_s$ , where the field is big enough to align all the Co magnetic moments. The resistance shows hysteresis when the magnetic field decreases. When the field direction is reversed, the resistance first increases, goes through a peak where the magnetic field is defined as  $H_p$ , and then decreases again to the minimum value at  $H_s$ . Decreasing and reversing the field again produces an almost, but not exactly, symmetric curve with respect to the field. The resistance curve then becomes stabilized.

Notice that the peak resistance is smaller than the as-deposited one. Resistance



Figure 3.2. (a) CPP resistance, (b) CIP resistance, and (c) magnetic moment versus field for a  $[Co(6nm)/Ag(6nm)] \times 60$  sample. CPP area  $A = 1.29 \text{ mm}^2 \pm 2.9\%$ 

.



Figure 3.3. (a) CPP resistance, (b) CIP resistance, (c) magnetic moment versus field for a  $[Co(50nm)/Ag(50nm)] \times 7$  sample. CPP area  $A = 1.20 \text{ mm}^2 \pm 4.2\%$ 

.



Figure 3.4. (a) CPP MR, (b) CIP MR versus field of a  $[Co(6nm)/Ag(6nm)] \times 60$  sample. (c) CPP MR of a Nb/Co(9nm)/Nb sandwich.

differences at  $H_0$  and  $H_p$  are believed to be results of different magnetic structures. However, the detailed magnetic structures of the as-deposited samples and at the  $H_p$ are not known. We have tried to bring the resistance back to the  $H_0$  value but were not usually successful, see section 5.1.6 for more details. On the other hand, if we sputter nominally the same sample again, the data, including the as-deposited value, are reasonably reproducible (see section 3.2.3).

The magnetic moment curves show hysteresis similar to bulk ferromagnetic material. The as-deposited sample usually has a magnetic moment close to zero. The peak positions of the CPP and CIP resistances are slightly different, close to, but usually larger than the coercive field  $H_c$ , where the magnetic moment goes through zero. Most of our samples are well saturated at 1kOe, a relatively small field compared with anti-ferromagnetically coupled Fe/Cr and Co/Cu systems which need approximately 20kOe and 10kOe, respectively, to be saturated. These lower fields, plus lack of coupling oscillation in other systems when spacer layer thickness is larger than 6nm, lead us to believe that Co/Ag multilayers are (partially) ferromagnetically coupled for Ag thicknesses less than 6nm and magnetically uncoupled for Ag thicknesses equal to or greater than 6nm, see section 5.1.3.

Samples with Co layers thicker than 12nm sometimes show more complex behavior. The as-deposited sample can have a resistance comparable to or smaller than that at the peak. Steps in the resistance and magnetization curves have been seen on some samples. Figure 3.3 shows an example. In (a) it's hard to tell if there is a step on the negative field side, because of large error bars (shown on the peak and at saturation) due to noise at the SQUID output. But for the peak on the positive field side, where the error bars are comparable to the symbol size, it is clear that there are steps and shoulders. In (b) and (c), where the error bars are about the symbol sizes, structures are clearly seen. The upper curve in (c) was measured twice to confirm the data. Comparing the three graphs, we see that the CPP resistance peak occurs at one of the magnetic moment steps and the CIP resistance peak occurs at another. To understand these thick Co layer samples one would have to study the detailed magnetic structure. We shall not focus on this issue here.

#### 3.2.2 Magneto-resistance

We define the magneto-resistance (MR) at field H as

$$MR(H) = \frac{R(H) - R(H_s)}{R(H_s)}$$

,

or with R replaced by AR for CPP geometry. We choose the saturation resistance as our reference point because that is the only state for which we know the magnetic structure, i.e., all the Co domains are aligned with the external field. This definition gives a larger MR than the definition with the zero field resistance chosen as the reference. Care must be taken when comparing MRs with different definitions. The Nb/Co interface resistance in the CPP geometry has a non-zero contribution in the denominator. Thus all our CPP MR data are smaller than the true MR of the multilayers. We shall keep this in mind in our general analysis, see section 5.1.

Figure 3.4 parts (a) and (b) replot Figure 3.2 (a) and (b) as MR versus field. Part (c) shows the CPP MR of a Nb/Co(9nm)/Nb sandwich on the same scale as (a) to show that neither the Nb/Co interfaces nor single Co layers make significant contributions to the MR. The CPP MR is more than four times larger than the CIP MR for this sample. Zhang and Levy [111] predicted that the CPP MR should be significantly larger than the CIP MR. We confirm this prediction for our Co/Ag samples. For a complete comparison between the CPP and CIP MR, see chapter 5.

#### 3.2.3 Reproducibility

We check the reproducibility of the CPP AR, CPP and CIP MR, and MR peak position and coercive field of  $[Co(6nm)/Ag(6nm)] \times 60$  multilayers.

#### **CPP** AR

We concentrate on three different fields,  $H_0$ ,  $H_p$ , and  $H_s$ , to check the reproducibility of the CPP AR. Table 3.1 shows six nominally identical  $[Co(6nm)/Ag(6nm)] \times 60$ samples. The column 'remark' lists the sample fabrication procedure, as discussed in section 2.2. No significant difference can be seen between procedures A and B. The average and standard deviation  $(\sigma_{n-1})$  of the ARs at different fields, in  $f\Omega m^2$ , for these samples are:

$$AR(H_0) = 120.9 \pm 3.6$$
  
 $AR(H_p) = 94.5 \pm 2.6$   
 $AR(H_s) = 68.3 \pm 3.2$ 

The standard deviations are 3-5% of the average value. The errors listed in the table are random errors, dominated by the area measurement. For a given sample, the ARs at different fields share the same error.

#### **CPP** MR

A sample with a larger resistance at  $H_0$  does not necessarily have a larger resistance at  $H_s$ . This lack of correlation leads to fluctuations in the MR data. Table 3.2 lists CPP MR data for these six samples. The errors listed in the table are small because we can measure resistances very accurately. The average values are:

CPP MR(
$$H_0$$
) = 77.3 ± 7.7

CPP MR(
$$H_p$$
) = 38.6 ± 3.9

The standard deviations are 10% of the average values for CPP MR data.

#### **CIP MR**

The CIP MR data of these samples are listed in Table 3.3. The top half of the table lists those samples on which the CIP MR were measured simultaneously with the CPP MR, with the current perpendicular to the field. The bottom half of the table lists data from simple square film samples, with the current parallel to the field. The two samples with Nb buffer layers were measured at temperature 10K so that the Nb was not superconducting. We tried different substrates, and growing multilayers on a Nb buffer layer, to see if these changes had a significant effect on the MR. We see no systematic change for samples on sapphire or Si substrate, with or without a Nb buffer, for different field directions. If we average over all of the samples, we find:

CIP MR( $H_0$ ) = 19.7 ± 4.1 CIP MR( $H_p$ ) = 11.5 ± 1.8

The standard deviations are 21% and 16% of the average values, respectively.

#### **Peak Position and Coercive Field**

The  $H_p$  positions of the CPP MR and CIP MR on the same sample, and the CIP MR on the corresponding single film are listed together with the coercive field of the single film in Table 3.4. We see that, the coercive fields of our multilayer samples all lie within the MPMS measuring uncertainty. However, the peak fields of CPP and CIP MR vary over much wider field ranges. The magnetization results suggest that the magnetic structures of our Co layers are similar, but the MR depends on not

only these magnetic structure but also on the sample physical structure. In general,  $H_p(\text{CPP}) \approx H_p(\text{CIP}) > H_c$  for these samples.

To conclude, our CPP AR measurements are reproduced within  $\pm 5\%$  of average values. When we calculate the MRs, however, both CPP and CIP data have much bigger fluctuations,  $\pm 10-20\%$ . The positions of the MR peaks also spread over wider ranges than do the coercive fields. Apparently the MR, which involves resistances at two fields, is very sensitive to the detailed sample structure. The detailed structures of the samples need to be studied to understand the sources of the scatter in the MR data. We will analyze CPP AR data quantitatively, but MR data only qualitatively.

|   | sample no. | A              | R(fΩm | <sup>2</sup> ) | error % | remark <sup>a</sup> |
|---|------------|----------------|-------|----------------|---------|---------------------|
|   |            | H <sub>0</sub> | $H_p$ | H,             |         |                     |
| 1 | 246-01     | 113.9          | 95.4  | 67.2           | 2.9 %   | Α                   |
| 2 | 298-01     | 123.9          | 98.8  | 74.6           | 3.2 %   | Α                   |
| 3 | 310-01     | 121.8          | 92.7  | 65.6           | 2.3 %   | Α                   |
| 4 | 336-01     | 123.2          | 91.4  | 67.6           | 2.8 %   | В                   |
| 5 | 337-01     | 120.3          | 94.7  | 67.1           | 3.3 %   | В                   |
| 6 | 339-01     | 122.0          | 93.7  | 67.4           | 1.8 %   | В                   |

Table 3.1. AR values at three magnetic fields of six different  $[Co(6nm)/Ag(6nm)] \times$  60 samples.

<sup>a</sup>Sample fabrication procedure. See section 2.2.

Table 3.2. CPP MR values at magnetic fields  $H_0$  and  $H_p$  of six different  $[Co(6nm)/Ag(6nm)] \times 60$  samples.

|   | sample no. | CPP MR% |       | error     |
|---|------------|---------|-------|-----------|
|   |            | $H_0$   | $H_p$ |           |
| 1 | 246-01     | 69.6    | 42.0  | $\pm 0.5$ |
| 2 | 298-01     | 66.0    | 32.4  | ±0.8      |
| 3 | 310-01     | 85.6    | 41.2  | ±0.3      |
| 4 | 336-01     | 82.4    | 35.3  | ±0.2      |
| 5 | 337-01     | 79.2    | 41.4  | ±0.2      |
| 6 | 339-01     | 81.0    | 39.0  | ±0.2      |

Table 3.3. CIP MR values at magnetic fields  $H_0$  and  $H_p$  of  $[Co(6nm)/Ag(6nm)] \times 60$  samples. The top half is data for those samples on which the CIP MR were measured simultaneously with the CPP MR. The bottom half is data for simple square films.

|    | sample no. | CIP MR%          |       | error     | substrate | remark         |
|----|------------|------------------|-------|-----------|-----------|----------------|
|    |            | $H_0$            | $H_p$ |           |           |                |
| 1  | 246-01     | 14.5             | 9.1   | ±0.5      | sapphire  | badª           |
| 2  | 298-01     | 19.0             | 8.7   | $\pm 0.5$ | sapphire  |                |
| 3  | 310-01     | 7.3              | 4.4   | ±0.5      | sapphire  | bad            |
| 4  | 336-01     | 20.6             | 9.0   | ±0.5      | sapphire  | bad            |
| 5  | 337-01     | 25.5             | 10.9  | $\pm 0.5$ | sapphire  |                |
| 6  | 339-01     | 25.8             | 11.5  | $\pm 0.5$ | sapphire  |                |
| 7  | 246-04     | N/A <sup>b</sup> | 11.2  | $\pm 0.5$ | sapphire  |                |
| 8  | 246-02     | $N/A^{b}$        | 10.7  | $\pm 0.5$ | Si        |                |
| 9  | 246-03     | N/A <sup>b</sup> | 11.6  | $\pm 0.5$ | Si        |                |
| 10 | 260-06     | 17.4             | 16.0  | $\pm 0.5$ | Si        | 40nm Nb buffer |
| 11 | 260-07     | 16.7             | 12.6  | $\pm 0.5$ | Si        | 20nm Nb buffer |
| 12 | 298-06     | 14.1             | 11.5  | $\pm 0.5$ | Si        |                |
| 13 | 308-07     | 19.4             | 10.5  | $\pm 0.5$ | Si        |                |
| 14 | 310-06     | 19.9             | 11.6  | $\pm 0.5$ | Si        |                |

<sup>a</sup>Misaligned CIP sample (see section 2.2).

<sup>b</sup>We didn't appreciate the irreversibility of  $H_0$  value for the first a few samples. Samples were saturated before any measurements were made.

Table 3.4. Magnetic fields  $H_p$  and  $H_c$  of  $[Co(6nm)/Ag(6nm)] \times 60$  samples.

| sample $H_p(Oe)$ |     | 0e) | $H_c(Oe)$ | error    | substrate | remark         |
|------------------|-----|-----|-----------|----------|-----------|----------------|
| no.              | CPP | CIP |           |          |           |                |
| 246-01           | 86  | 74  |           | $\pm 15$ | sapphire  | bad            |
| 298-01           | 103 | 91  |           | ±15      | sapphire  |                |
| 310-01           | 74  | 69  |           | ±15      | sapphire  | bad            |
| 336-01           | 74  | 65  |           | ±15      | sapphire  | bad            |
| 337-01           | 112 | 91  |           | $\pm 15$ | sapphire  |                |
| 339-01           | 86  | 74  |           | ±15      | sapphire  |                |
| 246-04           |     | 80  | 60        | ±10      | sapphire  |                |
| 246-02           |     | 80  | 60        | ±10      | Si        |                |
| 246-03           |     | 85  | 60        | ±10      | Si        |                |
| 260-06           |     | 80  | 65        | ±10      | Si        | 40nm Nb buffer |
| 260-07           |     | 85  | 70        | ±10      | Si        | 20nm Nb buffer |
| 298-06           |     | 90  | 75        | ±10      | Si        |                |
| 308-07           |     | 90  | 70        | ±10      | Si        |                |
| 310-06           |     | 80  | 65        | ±10      | Si        |                |

# **CHAPTER 4**

# Theory

In this chapter we first look at ferromagnetism in terms of itinerant electron theory. Then we introduce the current development of classical and quantum models for the transport properties of magnetic multilayers in both the CIP and CPP geometries.

# 4.1 Magnetism

#### 4.1.1 Ferromagnetism, bulk transition metals

Before going into multilayers, we first summarize briefly the most useful model of why bulk transition metals such as Fe, Co, and Ni are ferromagnetic.

The magnetic moment of transition metals is associated with the spin of electrons occupying partially filled d atomic orbits. Electron spin has only two possible projections, which we can take to be "up" and "down" with respect to a certain axis. Each electron then contributes to the macroscopic magnetization an elementary magnetic moment of one Bohr magneton ( $\mu_B$ ). For example, an isolated Co atom has outer shell electron configuration  $3d^74s^2$ . According to Hund's Rules, five electrons in the 3d shell of cobalt have their spins all pointing in the same direction and the remaining two are oriented in the opposite direction. Thus an isolated Co atom has a magnetic
moment of  $3\mu_B$  associated with the spin of 3d electrons. For isolated atoms, there is also a contribution of the orbital angular momentum of electrons to the total magnetic moment but since the orbital moment is quenched in a crystal we can neglect this term [109].

When atoms are brought close together to form a crystal, electron orbits overlap and electrons can move from atom to atom. Two electrons with parallel spins cannot occupy the same orbit in an atom, according to the Pauli exclusion principle, thus they do not experience strong Coulomb repulsion. On the other hand, whenever two electrons with opposite spins come on the same atom the energy increases by an amount equal to the intra-atomic Coulomb repulsion, U, assumed to be constant for simplicity. To minimize the total energy, it is advantageous for all d electrons to have their spins parallel, i.e. intra-atomic Coulomb repulsion favors formation of a spontaneous magnetic moment (ferromagnetic ordering) in the crystal. However, putting all electrons into states with the same spin orientation increases their total kinetic energy because advantage is not being taken of the possibility to occupy each state with two electrons. We thus have two competing tendencies that have to be balanced to decide whether ferromagnetic ordering takes place.

The densities of states  $D_s(E)$  and  $D_d(E)$  in the s and d bands of a nonmagnetic transition metal are shown schematically in Figure 4.1(a) for both spin orientations  $(\uparrow, \downarrow)$ . The area under each D(E) curve is equal to the total number of states of a given spin available to electrons. This is N for s electrons and 5N for d electrons, where N is the number of atoms in the crystal. Since the d band is narrow (d electrons are sluggish) and must accommodate five times as many electrons as the s band, we have  $D_d(E) \gg D_s(E)$ . The bands depicted in Figure 4.1(a) are for a metal with zero total magnetic moment.

Transferring electrons from one spin band to the other increases their band (kinetic) energy. Consider a metal with a surplus of d electrons in the up spin band,



Density of states

Figure 4.1. (a) Schematic representation of the densities of states in the s and d bands of a non-magnetic metal at absolute zero. (b) Schematic representation of the densities of states in the s and d bands of ferromagnetic cobalt at absolute zero. The bands are filled up to the common Fermi level  $E_F$ . (After reference [113]).

 $n_1 > n_1$ . We know that the kinetic energy is increased by this imbalance but the Coulomb interaction energy is reduced by  $(n_1 - n_1)U$ . The reduction in the interaction energy causes a downward shift of the up spin band by  $(n_1 - n_1)U$ . If U and  $D_d(E)$  are large, the reduction in the interaction (potential) energy may outweigh the increase in the kinetic energy and a ferromagnetic state  $n_1 > n_1$  may become stable. In such a case the up spin band shifts downward, and the two spin bands become split. The condition for this to occur is the Stoner condition  $\frac{U}{N}D_d(E_F) > 1$  [114]. Because the density of states is small for an s band, the Stoner condition is not satisfied and the up and down spin bands of s electrons are not split. The Stoner condition is satisfied for d-bands in Fe, Co, and Ni. The equilibrium shifts of the two spin bands for Co are shown schematically in Figure 4.1(b) (after reference [113]).

For more detailed discussions about magnetism, see Stoner [114], Herring [115], and Jiles [116].

### 4.1.2 Magnetism in Multilayers

As mentioned in Chapter 1, when certain ferromagnetic and non-ferromagnetic thin films are deposited alternately to form sandwiches or multilayers, the alignment of magnetic moments of neighboring ferromagnetic layers is found to oscillate between parallel and antiparallel depending on the thickness of the non-ferromagnetic layer. These spontaneous alignments mean that there are magnetic couplings between the ferromagnetic layers mediated by the spacer layers. Experimental results show: (1) For certain crystal orientations, and for sufficiently good samples, there is a short period coupling superimposed on a long one. For samples with poorer quality, e.g. rougher interfaces, the short period coupling tends to be smeared out. The long period coupling depends less on the sample quality. (2) The coupling strength, or the amplitude of the oscillation, diminishes with increasing spacer layer thickness. Since the giant magnetoresistance effect is ascribed to the interplay between spin dependent scattering in successive magnetic layers, it is important to know the coupling in these magnetic structures.

The detailed theory of the magnetic coupling in magnetic multilayers is out of the scope of this dissertation. Only a brief survey will be presented of the principles and results of the two basic strategies that have been used to study the interlayer exchange coupling: total energy calculations and perturbative models.

### **Total Energy Calculations**

Total energy calculations are in principle straightforward: the coupling is calculated as the energy difference between the states with parallel and antiparallel magnetization alignments. Such calculations have been performed either from first principles [117, 118] or within a tight-binding scheme [119]. This kind of calculation is very difficult because the energy difference is several orders of magnitude smaller than the total energy. Thus, theorists have to improve their computer programs to get better numerical accuracy. Total energy calculations are also very demanding in computer time because the magnetic unit cell must be twice the chemical unit cell and the computation time increases with the size of the unit cell. For the same reason, it is hard to study the effect of interface roughness on the interlayer coupling.

A lot of materials and crystalline orientations have been studied. Coupling between ferromagnetic layers oscillating with a period of about two-monolayer-thick spacer layers was found in some of these calculations. This finding agrees with some refined experiments, but the calculated coupling strength is much too large. When the interface was modeled as ordered compounds of the two constituents, the short period oscillation was damped and even disappeared (see e.g.[54] for a review). To our knowledge, no long-period coupling oscillations have been obtained by total energy calculations.

#### **Alternative Approach**

An alternative approach to calculate the interlayer coupling is to obtain the coupling directly without computing the total energy. The price to pay is one has to make approximations suggested by physical intuition.

A number of different models have been proposed [111-122]. They all rely on the same underlying picture for the coupling mechanism. The ferromagnetic layer spin polarizes the conduction electrons of the spacer, this spin-polarization extends through the spacer and interacts with neighboring ferromagnetic layers, thus giving rise to an effective exchange interaction. The various models differ mostly in the assumptions about the physical system and in the simplifying approximations made. Nevertheless, it is a common feature of all of them that when the spacer thickness, z, is about the order of a couple monolayers thick, the coupling oscillates periodically with an oscillation period related to some measure of the Fermi surface of the spacer metal, and with an amplitude decaying like  $1/z^2$ , see below. **RKKY Model** The Ruderman-Kittel-Kasuya-Yosida (RKKY) model is the archetype of the perturbative theories of interlayer coupling. With an assumed spherical Fermi surface for the interlayer materials with Fermi wave vector  $k_F$ , and when the interlayer thickness is large relative to its lattice parameter, the RKKY theory predicts that the interlayer coupling oscillates with a period  $\lambda = \pi/k_F$  and the amplitude decreases with  $z^2$ . For spherical Fermi surfaces, this period is about one monolayer, which is too short even compared with the shorter period in refined experiments. This discrepancy is reconciled by taking into account that the spacer layer thickness does not vary continuously, but must be an integer multiple of monolayer thickness d. Because of this discrete sampling on the interlayer coupling, the effective period  $\Lambda$  is given by

$$\frac{1}{\Lambda} = \left| \frac{1}{\lambda} - \frac{n}{d} \right|$$

where the non-negative n is chosen such that  $\Lambda \ge 2d$ . This effect is called *aliasing* [123, 128].

The multi-period oscillations observed experimentally [28, 29] in certain structural orientations can also be explained by performing the calculation without approximating the F layer as continuous, and by taking into account the real Fermi surfaces. In short, each Fermi surface spanning vector, which points in a direction that has corresponding Fermi velocities antiparallel to one another, contributes an oscillatory coupling period [122].

To summarize, models based on RKKY theory for realistic Fermi surfaces are able to explain the observed periods of oscillatory coupling correctly. However, they are not able to describe correctly the strength or the phase of the coupling oscillations [120, 122, 123].

**Other Models** Other models explain the exchange coupling in terms of a confinement of d-holes in the spacer layers [126] and exploit an analogy between the

coupling and the de Haas-van Alphen effect; or use quantum-well states to analyze the spacer layer electronic structure in the framework of states consisting of bulk Bloch functions modulated by an envelope function [131]. The predictions from these models have similar form to those of the RKKY models.

Models taking into account the hybridization between the 3d bands of the ferromagnetic layer and the conduction electrons of the spacer layer are also proposed to describe experiment results. The interaction can be broken down into an RKKY-type contribution and a superexchange-type [132] contribution. The superexchange interaction comes from virtual excitation processes that are not tied to the Fermi surface. It was shown to be essential to include this additional coupling to explain refined experiment data on Fe/Cr [124, 125].

For a review of different models for magnetic coupling in multilayers, see reference [54].

## **4.2** Electron Transport

### 4.2.1 Band Theory

### **Transition** Metals

At low temperatures, electrons in metals are scattered mainly from impurities. Since scattering from impurities is elastic (the energy is conserved), and all the states with energy lower than the Fermi energy  $E_F$  by more than  $k_BT$  are occupied, electrons can be scattered only to states near the Fermi level. Thus the scattering probability is proportional to the density of states at  $E_F$ ,  $D(E_F)$ .

To oversimplify somewhat, in the case of noble metals such as silver, with electron configuration  $4d^{10}5s^1$ , the Fermi level intersects only an s band and the density of states in the s band is low. Thus the scattering probability is low and noble metals

are very good conductors.

The Fermi level in transition metals, in contrast, intersects both s and d bands. The density of states in the d band is very high, which opens up a new effective channel for scattering of conduction electrons into the d band. This scattering mechanism (Mott scattering [133]) explains why transition metals are poor conductors compared with noble metals.

The conduction electrons in metals are usually considered not spin-polarized. However, it has been pointed out that conduction bands in transition metals have some spin-polarized character [134, 135].

There are electron scattering processes in which the orientation of the spin changes (spin flip scattering), e.g. spin-orbit scattering, or electron magnon scattering in ferromagnetic metals. At low temperature, the spin flip scattering of conduction electrons by magnons is frozen out and the residual spin flip scattering is due to spin orbit interaction. This latter process has been extensively studied by Electron Spin Resonance and it is known experimentally that the probability is very small in pure metals [136, 137]. Therefore, it is justified to assume that the current carried by up spin electrons is independent of the current carried by down spin electrons when no strong spin-orbit scattering center is present. In a transition metal ferromagnet such as Co, the up and down spin d bands are split and the densities of states at the Fermi level seen by up and down spin conduction electrons are very different. It follows that the mean free path in a ferromagnet is spin dependent and one has to define separate resistivities for up and down spin electrons [71].

### **Magnetic** Multilayers

The actual band structures for magnetic multilayers are very complicated because of the ferromagnetic/non-magnetic layering. If one focuses on conduction from an unpolarized band of electrons, s-electrons, the MR effect comes from the spin-dependent scattering. However, conduction electrons in transition metal multilayered structures have spin-polarized character. In the Stoner description of conduction electrons in ferromagnetic layers, one needs to consider different potentials for majority and minority spins. When electrons go from a ferromagnetic layer into a paramagnetic layer, or into another magnetic layer which has different magnetization direction, they will be partially transmitted or scattered. The role of spin-polarized conduction electrons in the magnetic multilayers was not examined until recently [135, 138], we will discuss the results later in this chapter.

### 4.2.2 Boltzmann and Kubo Approach

Two basic approaches are used to describe electron transport in magnetic multilayers. One starts from the semiclassical Boltzmann equation and the other from the Kubo linear response formalism, which is a quantum theory. The underlying physics of the Boltzmann equation is to describe how the perturbation of the electron distribution function by a given magnetic layer spreads and interacts with the perturbation produced by neighboring magnetic layers. The Kubo formalism, together with assumed potentials and scatterers (see below) in magnetic multilayers, is used to solve for the Green's function involved and to determine the conductivity of the structure.

There are two scattering potential contributions when describing conduction electrons in magnetic multilayers. One is the "spin-dependent superlattice potential" of the electrons due to the different layers and the other is the "spin-dependent scatterer", e.g. impurity sites within the bulk of the layers and at the interfaces. Most attention has been focused on the role of the inhomogeneous spin-dependent scatterer, i.e. different distribution of scattering centers from one layer to the next, in producing large MR in multilayers. The scattering has been evaluated by using plane wave states. Recently, models incorporating the spin-dependent superlattice potential, in which more appropriate Bloch-like wavefunctions are used, were studied by various groups [135, 138, 139].

### 4.2.3 CIP MR versus CPP MR

The current density, j, and electric field, E, are related by

$$j = \sigma \cdot E \quad . \tag{4.1}$$

For the CIP geometry, the electric field is uniform throughout the multilayer, and the current varies from one layer to another as well as within each layer. For the CPP geometry the current is constant throughout, while the electric field varies from layer to layer. Therefore, theorists calculate conductivities in CIP, but resistivities, the inverse of conductivities, in CPP geometry. In a two channel model, each spin channel is treated separately and the total resistivity/conductivity is obtained by combining the two channels in parallel.

### **CIP MR**

The discovery of giant MR in magnetic multilayers was made by Baibich *et al.*, who gave a qualitative interpretation for their experimental results in Fe/Cr superlattices along the following lines [17]: (1) Perfect interfaces produce only specular reflection and diffraction of the electron waves; these processes do not change the resistivity significantly. It is scattering from interface roughness that affects the resistivity. (2) The giant MR results from spin-dependent transmission of the conduction electrons between ferromagnetic layers, which are in antiferromagnetic configuration at zero field and in ferromagnetic configuration above a saturation field. When the superlattice is in its ferromagnetic configuration, the alignment of the magnetizations opens up one spin channel and lowers the resistivity. From then on, different models based upon this simple picture were proposed to describe giant MR. In all the models for the CIP MR, the elastic mean free paths (MFP) of the spin-up and spin-down electrons are the dominant characteristic lengths. In the limit of long MFP, electrons travel through several magnetic layers and interfaces before being scattered inelastically, thus sampling a large part of the multilayer. In the short MFP limit, multilayers become independent layers as far as the electrons are concerned, thus the interplay between different layers vanishes. The MR dependence on the thicknesses of individual layers can be predicted for extremely short or long mean free paths and predictions can be compared with experimental results. It is a common feature of all the models for the CIP MR that either a lot of assumptions and simplifications have to be made to investigate the MR dependence on certain parameters or, in the most general cases, many adjustable parameters are needed to describe experimental results.

**Boltzmann Equation** Camley and Barnas [64] extended to *magnetic* multilayers the theory used to describe conduction in thin films by Fuchs and Sondheimer [140] and extended to multilayers by Carcia and Suna [141]. They solved the Boltzmann transport equation in real space in each layer and matched the solutions at the interfaces. They assumed spin-dependent scattering both within the bulk ferromagnet and at the interfaces and only transmission or diffusive scattering at the interfaces. To take into account the momentum transfer between two electron spin channels when the ferromagnetic layers are not aligned, a transmission coefficient was introduced at the center of the non-magnetic layer. Other assumptions include free electron gases, the same Fermi energies, and the same elastic MFP values for both metals. With all these simplifications, this model was able to qualitatively describe the main features of experimental data using only two parameters: the diffusive scattering parameter for one spin channel and the ratio of this scattering between the two spin channels. The model of Camley and Barnas has been extensively used for numerical calculations of the MR in sandwiches and multilayers [64, 76, 142]. The same semi-classical approach has also been used by different groups to derive analytical expressions for MR in some simple limits. Barthelemy and Fert [143] derived expressions in the limit where magnetic layer thickness is much larger than the MFP. Edwards *et al.* [144] used a slightly different approach to develop a resistor network model and derived expressions for MR in the limit of very long and very short MFP compared to layer thicknesses.

Some improvements have been introduced in the model of Camley and Barnas. To improve the treatment of transmission and diffuse scattering coefficients, Johnson and Camley [80] introduced interfacial regions modeled as additional layers. Dieny [145, 146] took into account the granular structure of sputtered samples by introducing anisotropy in the MFP. Dieny *et al.* also made a strong argument that the change in sheet conductance is the fundamental measure of the CIP MR [68], rather than the change in resistance, resistivity, or the ratio between the change in resistance (conductance) and total resistance (conductance), as almost all other groups had been calculating.

Hood and Falicov [135] updated the model of Camley and Barnas by taking into account the effect of the spin-dependent superlattice potential on the electron wave function (instead of free electron gas). They found that up to 20 parameters are needed to describe the MR behavior in the most general case.

**Kubo Formalism** In the quantum model of Levy and coworkers [147, 148, 149], the assumptions include: (1) free electron gas, (2) electrical current carried in parallel by up and down spin (majority and minority) electrons, (3) spin-dependent scattering within magnetic layers and at the interfaces. Using a Hamiltonian consisting of kinetic energy for free electrons and spin-dependent potential for scatterers described above, they solved for the Green's function in reciprocal (momentum) space and derived a position-dependent conductivity for layered structures. The predictions from their model include: (1) for interfacial spin-dependent scattering only, the MR ratio is a continuously decreasing function of both the magnetic and non-magnetic layer thicknesses, (2) for bulk spin-dependent scattering only, the MR variation with the thicknesses is more complex. The dependence of the MR on other parameters like the spin asymmetry etc. is described by Zhang *et al.* [148].

Zhang and Levy recently introduced a superlattice potential into their model [150]. The spin-dependent potential was described by a Kronig-Penny potential in the direction perpendicular to the layer plane and a constant in the plane of the layers. Their aim was to assess the relative importance of the superlattice potential versus spin-dependent scattering in producing the large MR. They found that a superlattice potential does not by itself give rise to a giant MR effect in the CIP geometry, but can reinforce or undermine the contribution from spin-dependent scattering.

Vedyayev et al. [151] also used the Kubo formalism to calculate the exact Green's function, under the assumption of free electrons in magnetic multilayers, in real space for the case of an infinite multilayer. Under the same assumptions as Camley and Barnas, they found reasonable agreement between Boltzmann and Kubo approaches [152].

Maekawa and co-workers [153] recently proposed a tight-binding model perpendicular to layer planes for a superlattice potential. They then solved for the Green's function in the Kubo formalism in real space. Their findings are similar to those of Zhang and Levy's [150].

### CPP MR

Zhang and Levy extended the quantum model they had developed for the CIP MR [147] and applied it to CPP MR [111]. Assuming there is no spin flip scattering, they calculated the CPP conductivity of multilayered structures in the local limit where the mean free path is small compared with individual layer thickness. They got a simple form for the CPP conductivity in which the conduction electrons are scattered by every interface and every layer, and which leads directly to a series resistance model for each spin channel. The total resistance of the structure is then obtained by combining the two spin channels in parallel. They found that the local resistivity is position dependent and has a length scale set by the mean free path, but this length is removed while calculating the total resistance. With this model, and using the best fit parameters from Fe/Cr CIP MR data, they predicted that the CPP MR should be much larger than the CIP MR.

Our group, following the ideas of Levy and co-workers [111], and Mathon and co-workers [144], first proposed a simple series resistance model without separating different spin directions to explain our data on Co/Ag, which it fit very well (see Lee *et al.* [154]). Since, in the CPP geometry, the measuring current has a uniform density and flows sequentially through each layer and interface, we assumed that the total resistance of the multilayer has three contributions: (a) one proportional to the non-magnetic (N) metal thickness,  $t_N$ , which we designate  $\rho_N$ ; (b) one field-dependent component proportional to the magnetic (F) layer thickness,  $t_F$ , which we designate  $\rho_F(H)$ ; (c) one field-dependent component proportional to the bilayer number, M, which we designate  $R_{F/N}(H)$ . For a sample as described in section 2.2, we write the area A times the total resistance  $R_T$  as

$$AR_{T}(H) = 2AR_{S/F} + M\rho_{F}(H)t_{F} + (M-1)\rho_{N}t_{N} + 2(M-1)AR_{F/N}(H)$$
(4.2)

where  $R_{S/F}$  is the superconductor-ferromagnetic interface resistance. M - 1 appears because when there is an integer number of bilayers in a sample, the topmost Ag layer has zero resistance(see page 36). If there is an extra Co layer in the sample, we can still use M if we redefine it to denote the number of Co layers. We showed [154] that our three sets experimental data for AR versus M at  $H_0$ ,  $H_p$ , and  $H_s$  (see Figure 3.2) are all consistent with straight lines as predicted by equation 4.2. They are Co/Ag samples (1) two sets with  $t_{Co} = \text{const.}$ ; (2) one set with  $t_{Co} = t_{Ag}$ , all with the total thicknesses  $t_T = Mt_{Co} + Mt_{Ag}$  fixed. However, subsequent Co/AgSn data showed clear curvature on a  $AR_T(H_0) - AR_T(H_s)$  versus M plot [155], which is inconsistent with this one-channel model. We therefore turned to a two-channel model to explain these Co/AgSn data.

Professor Pratt of our group then derived important equations in this model for the CPP geometry, see Lee *et al.* [155]. The assumptions made include (1) no spinflip scattering; (2) electrical current carried in parallel by up (+) and down (-) spin (majority and minority) electrons; (3) equal numbers of (+) and (-) current carrying electrons; (4)  $R_{S/F}$  is the same for (+) and (-) conduction electrons. We define:

$$\rho_F^{\dagger} = \frac{2\rho_F}{1+\beta} \tag{4.3}$$

$$\rho_F^{\downarrow} = \frac{2\rho_F}{1-\beta} \tag{4.4}$$

$$R_{F/N}^{\dagger} = \frac{2R_{F/N}}{1+\gamma} \tag{4.5}$$

$$R_{F/N}^{\downarrow} = \frac{2R_{F/N}}{1-\gamma} \tag{4.6}$$

$$\rho_N^{\dagger} = \rho_N^{\downarrow} = 2\rho_N \tag{4.7}$$

$$R_{S/F}^{\dagger} = R_{S/F}^{\downarrow} = 2R_{S/F} \quad . \tag{4.8}$$

 $\rho_F^{\rm I}$  is the F layer resistivity when the electron spin and the local moment  $\mathcal{M}_i$  are parallel to each other, and  $\rho_F^{\rm I}$  is the resistivity when they are antiparallel. Here  $\rho_F$ is the F resistivity measured on independent thin F films or S/F/S sandwiches [107]. Similarly,  $R_{F/N}^{\dagger}$  and  $R_{F/N}^{\rm I}$  are the F/N interface resistances.

When the F layers in a magnetic multilayer are in the antiferromagnetic (af) state,

with equation (4.2) for (+) and (-) channels, we have:

$$\begin{aligned} AR_{T}^{(+)}(af) &= 2AR_{S/F}^{\dagger} + \frac{M}{2}\rho_{F}^{\dagger}t_{F} + \frac{M}{2}\rho_{F}^{\downarrow}t_{F} + \frac{(M-1)}{2}\rho_{N}^{\dagger}t_{N} + \frac{(M-1)}{2}\rho_{N}^{\dagger}t_{N} \\ &+ \frac{(M-1)}{2}2AR_{F/N}^{\dagger} + \frac{(M-1)}{2}2AR_{F/N}^{\dagger} \\ &= 4AR_{S/F} + \frac{2}{1-\beta^{2}}M\rho_{F}t_{F} + 2(M-1)\rho_{N}t_{N} + \frac{2}{1-\gamma^{2}}(M-1)2AR_{F/N} \\ &= AR_{T}^{(-)}(af) \quad . \end{aligned}$$

Since the resistances of these two channels add in parallel,  $AR_T$  is then just half of these equal values:

$$AR_T(af) = 2AR_{S/F} + M\rho_F^* t_F + (M-1)\rho_N t_N + (M-1)2AR_{F/N}^*$$
(4.9)

where  $\rho_F^* = \rho_F / (1 - \beta^2)$  and  $R_{F/N}^* = R_{F/N} / (1 - \gamma^2)$ .

Equation (4.9) can be verified experimentally. For a selected multilayer system,  $\rho_F$ (and  $\beta$ ),  $\rho_N$ , and  $R_{F/N}$  (and  $\gamma$ ) are fixed;  $t_F$ ,  $t_N$ , and M can be varied systematically. It is obvious from equation (4.9) that  $AR_T(af)$  should be linear with  $t_F$ ,  $t_N$ , or Mwhen the other two variables are kept fixed. We can also make a series of samples with the total thickness  $t_T$  fixed, thus putting a constraint on  $t_F$ ,  $t_N$ , and M. When  $t_T$  and  $t_F$  are kept fixed, substituting  $t_F = \frac{t_T}{M} - t_N$  into equation (4.9), we have:

$$AR_{T}(af) = 2AR_{S/F} + \rho_{F}^{*}t_{F} + (1 - \frac{1}{M})\rho_{N}t_{T} + (M - 1)\left[(\rho_{F}^{*} - \rho_{N})t_{F} + 2AR_{F/N}^{*}\right]$$

٠

When  $t_T$  is kept fixed and we set  $t_F = t_N$ , then  $t_F = t_N = \frac{t_T}{2M}$ . The two channel model gives:

$$AR_{T}(af) = 2AR_{S/F} + \frac{1}{2}(\rho_{F}^{*} + \rho_{N})t_{T} - \frac{1}{2M}\rho_{N}t_{T} + (M-1)2AR_{F/N}^{*}$$

These equations are also linear in M when M is big enough so that  $\frac{1}{M}$  terms can be ignored.

When the multilayer is in the ferromagnetic (f) state, the two channels behave differently.

$$\begin{aligned} AR^{(+)}(f) &= 2AR_{S/F}^{\dagger} + M\rho_F^{\dagger}t_F + (M-1)\rho_N^{\dagger}t_N + (M-1)2AR_{F/N}^{\dagger} \\ &= 4AR_{S/F} + 2(M-1)\rho_N t_N + \frac{2}{1+\beta}M\rho_F t_F + \frac{2}{1+\gamma}(M-1)2AR_{F/N} \\ &= 4AR_{S/F} + 2(M-1)\rho_N t_N + 2(1-\beta)M\rho_F^* t_F + 4(1-\gamma)(M-1)AR_{F/N}^* \\ AR^{(-)}(f) &= 4AR_{S/F} + 2(M-1)\rho_N t_N + 2(1+\beta)M\rho_F^* t_F + 4(1+\gamma)(M-1)AR_{F/N}^*. \end{aligned}$$

For ease of derivation, we keep  $\beta$ ,  $\gamma$  and put

$$a = 4AR_{S/F} + 2(M-1)\rho_N t_N$$
  

$$b = 2M\rho_F^* t_F$$
  

$$c = 4(M-1)AR_{F/N}^*$$
.

We have

$$AR^{(+)}(f) = a + b(1 - \beta) + c(1 - \gamma)$$
$$AR^{(-)}(f) = a + b(1 + \beta) + c(1 + \gamma)$$

Adding these channels in parallel, we get

$$\frac{[a+b(1+\beta)+c(1+\gamma)][a+b(1-\beta)+c(1-\gamma)]}{[a+b(1+\beta)+c(1+\gamma)]+[a+b(1-\beta)+c(1-\gamma)]} = \frac{[a+b+c+(\beta b+\gamma c)][a+b+c-(\beta b+\gamma c)]}{2[a+b+c]}$$

\*The derivation is more tedious if we do not substitute  $\rho_F/(1+\beta)$  with  $(1-\beta)\rho_F^*$  etc. as above.

$$= \frac{1}{2}[a+b+c] - \frac{\frac{1}{4}(\beta b+\gamma c)^2}{\frac{1}{2}[a+b+c]}$$

Noting that  $\frac{1}{2}[a+b+c]$  is just  $AR_T(af)$ , we have:

$$AR_T(f) = AR_T(af) - \frac{[\beta M \rho_F^* t_F + \gamma 2(M-1)AR_{F/N}^*]^2}{AR_T(af)} \quad . \tag{4.10}$$

Rearranging terms we get the important relationship:

$$A\sqrt{R_T(af)[R_T(af) - R_T(f)]} = \beta \rho_F^* t_F M + 2\gamma A R_{F/N}^* (M-1) \quad . \tag{4.11}$$

Again  $t_F$  and M are related to each other by  $t_T = M(t_F + t_N)$ . Thus, for example, for a set of samples with total thicknesses  $t_T$  fixed and  $t_F = t_N$  varying at the same time, we should rewrite (4.11) as:

$$A\sqrt{R_T(af)[R_T(af) - R_T(f)]} = \beta \rho_F^* \frac{t_T}{2} + 2\gamma A R_{F/N}^*(M-1) \quad . \tag{4.12}$$

The left hand sides of equations (4.11),(4.12) are combinations of quantities that can be measured in CPP MR experiments. The right hand sides of these equations are independent of  $\rho_N$ , and the parameters  $\beta$  and  $\gamma$  can be determined once  $\rho_F^*$  and  $R_{F/N}^*$ have been found from fitting to the  $AR_T(af)$  equations. Equations (4.9),(4.10),(4.11) will be shown to be compatible with our Co/Ag and Co/AgSn data.

The model described above is developed for antiferromagnetically aligned systems. It is not clear that it should be applicable to uncoupled systems like our samples with thick spacer layers. To explain our CPP MR data, Zhang and Levy [67] re-formulated their theory to compare the MR of two different states with total magnetization equal to zero — one is the AF coupled state and the other is a state with a superposition of statistically uncorrelated magnetic configurations. They found that the CPP MR is the same for these two states. In contrast, the CIP MR is diminished for the "uncorrelated" state relative to the AF state.

When a ferromagnet, which has spin dependent resistivity, is put next to a nonmagnetic metal whose resistivity is not spin dependent, we should see an effect at the interface which is introduced as follows. The concept of "spin coupled interface resistance" was previously introduced by Johnson and Silsbee [156] and independently by von Son et al. [157] to describe the electron transport through an interface between ferromagnetic and non-magnetic metals. If in the ferromagnet the current is spin polarized, there will be spin accumulation around the interface with the nonmagnetic metal. This spin accumulation gives rise to an extra potential drop  $\Delta V_{I}$ , proportional to the current density J,  $\Delta V_I = r_{SI}J$  where  $r_{SI}$  is the "spin coupled interface resistance". This effect does not appear in the CIP geometry because there is no spin accumulation. Johnson [158] pointed out that spin accumulation must be taken into account in the CPP geometry. However, the characteristic length of this spin coupled resistance is of the order of spin diffusion length,  $l_{sf}$ . Valet and Fert [159] showed that in the limit of small bilayer thicknesses  $\Lambda$ , i.e.  $\Lambda \ll l_{sf}$ , applicable to most experiments, the spin accumulation manifests itself as an oscillation in the current density and does not contribute to the total resistance in the CPP geometry.

Valet and Fert [159] proved that, in the limit of mean free path much smaller than spin diffusion length, i.e.  $\lambda \ll l_{sf}$ , a Boltzmann equation model reduces to a macroscopic model. The results of the macroscopic model for the CPP conduction in magnetic multilayers are simple in the limit where the layer thicknesses are much smaller than the spin diffusion length  $l_{sf}$ . In this limit, the total resistance of a multilayer can be calculated by the two channel series resistance model. They also showed that, as did Camblong *et al.* [160], the spin diffusion lengths, rather than the mean free paths, are the relevant quantities in the CPP geometry.

Zhang and Levy [138], Maekawa and co-workers [153], and Barnas and Fert [161]

recently took into account the superlattice potentials on the transport in magnetic multilayers. They found that the superlattice potential plays an important role in the CPP MR but is much less important in the CIP MR. In general, they concluded that both spin-dependent scattering and the effects of superlattice potentials are needed to explain the MR in multilayer structures.

To summarize, different models in electron transport have been developed for the CPP MR. It was predicted that the CPP MR is larger than the CIP MR; we shall show that this is generally true for our Co/Ag and Co/AgSn multilayers. A simple one channel model was used to describe our Co/Ag data. It worked well on the Co/Ag but could not explain the behavior of our Co/AgSn data. A two channel model describing the CPP resistances in both AF and F states was then developed to explain both our Co/Ag and Co/AgSn data. We shall show in chapter 5 that this model, equations (4.9) and (4.10) together with  $t_T = M(t_F + t_N)$ , is compatible with most of our data.

# CHAPTER 5

## **Data Analysis**

In pursuit of the dependence of the Co/Ag multilayer's resistance, magnetoresistance (MR), and magnetization on its constituents, we made different sets of samples with Co layer thickness, Ag layer thickness, repeat bilayer numbers M, and consequently total thickness as variables. These variables satisfy the equation

$$t_T = M(t_{Co} + t_{Ag})$$

for each individual sample, thus there are only three independent variables. We systematically kept two of these variables the same in each set of samples and changed one. In one set, 4 at.% Sn was added to the Ag to test the effect of reducing the electron mean free path of the normal metal. More study on alloy interlayers is ongoing.

We present eight sets of data:

- $[Co(6nm)/Ag(6nm)] \times M$  Co and Ag layer thicknesses are kept at 6 nm, bilayer number M is varied and thus the total sample thickness.
- [Co(6nm)/Ag(t)]×M Co layer thickness is kept at 6nm, Ag layer thickness is varied. Total sample thicknesses are kept as near as possible to 720nm with integer bilayer number, so the bilayer number changes with Ag layer thickness.

- [Co(2nm)/Ag(t)]×M Co layer thickness is kept at 2nm, Ag layer thickness is varied. Total sample thicknesses are kept as near as possible to 720nm with integer bilayer number.
- [Co(t)=Ag(t)]×M Co and Ag layer thicknesses are kept equal and varied at the same time. Total sample thicknesses are kept as near as possible to 720nm with integer bilayer number. Bilayer number varies with individual layer thickness.
- [Co(t)/Ag(6nm)]×M Ag layer thickness is kept at 6nm, Co layer thickness is varied. Total sample thicknesses are kept as near as possible to 720nm, with integer bilayer number.
- [Co(6nm)/Ag(t)]×60 Co layer thickness is kept at 6nm and bilayer number at 60, Ag layer thickness is varied and thus the total sample thickness.
- [Co(t)/Ag(6nm)]×60 Ag layer thickness is kept at 6nm and bilayer number at 60, Co layer thickness is varied and thus the total sample thickness.
- [Co(6nm)/AgSn(t)]×M Co layer thickness is kept at 6nm. Nominally 4 at.% of Sn was added to Ag to decrease the elastic mean free path. AgSn layer thickness is varied. Total sample thicknesses are kept as near as possible to 720nm with integer bilayer number.

Some samples are repeated in different groups, like  $[Co(6nm)/Ag(6nm)] \times 60$  samples. All these samples will be listed repeatedly in different categories where they fit.

Because this is a pioneer work in the field, we present all of our data for archival purposes. In sections 5.1 to 5.3, MR data for both the CPP and CIP geometries, and magnetization data, are discussed qualitatively. Then, in section 5.4 and 5.5, we present our data for the CPP resistance and show that the two channel series resistance model is compatible with our data. Quantitative analysis with this model has good success in describing most, but not all, of our data. We will speculate on the sources of observed deviations.

## 5.1 CPP and CIP MR

We define the magneto-resistance (MR) at field H as

$$MR(H) = \frac{R(H) - R(H_s)}{R(H_s)} , \qquad (5.1)$$

or with R replaced by AR for the CPP geometry. There is a "contact resistance" contribution,  $2AR_{Nb/Co}$ , in our measuring technique for the CPP geometry (see chapter 3). This term is magnetic field independent under the fields we used, because the Nb stays superconducting. In equation 5.1, this term is cancelled in the numerator but not in the denominator. Thus all our CPP MR data are smaller than the true MR of the multilayers. We shall note when our analysis is altered because of this "contact resistance".

There are three important questions we would like to answer by analyzing our MR data. First, is the CPP MR always larger than the CIP MR? This was shown to be true for  $[Co(6nm)/Ag(6nm)] \times 60$  samples in section 3.2. Whether it is true for different Co and Ag thicknesses still needs to be shown. Second, is there magnetic ordering in our samples in the thickness range we are studying? If the answer is yes, what is the dependence of the ordering on layer thicknesses? Third, what is the relative importance of bulk versus interface scattering on giant MR?

Since the MRs of our nominally identical samples can vary by  $\pm 10\%$  around the average value for the CPP geometry and by as much as  $\pm 20\%$  for the CIP geometry (see section 3.2.3), we analyze our data only qualitatively here.

### 5.1.1 Experimental Results

In Tables 5.1-5.8 we present CPP and CIP MR data for eight sets of samples in the order in which we will discuss the data. For a large number of the samples we made equivalent simple square films (see chapter 2.1 for details). The CIP MRs of such films are listed on the right side of the tables. For simplicity, only those few of our  $[Co(6nm)/Ag(6nm)] \times 60$  samples that have either the largest or smallest MRs are listed in the tables in this section. Complete values of all the  $[Co(6nm)/Ag(6nm)] \times 60$ samples are given in section 3.2.3.

### 5.1.2 CPP MR > CIP MR

From the tables, we see that the CPP MR is larger than the CIP MR for all of our Co/Ag and Co/AgSn samples. For our Co/Ag samples, the ratio between the CPP MR and the CIP MR,  $\Pi$ , at both  $H_0$  and  $H_p$ , ranges between 2.5 and 6 for reliable CIP MR. Samples with thicker Ag layers tend to have larger  $\Pi$ , while samples with thicker Co layers tend to have smaller  $\Pi$ . In reference [81] we published some data which showed  $\Pi \geq 10$ . We now categorize those as samples with misaligned CIP parts. The misalignment was not obvious at that time, but parts of the misaligned samples became discolored after being exposed to the atmosphere for a long time. We mark these samples in the tables as "bad".

Figures 5.1 shows two sets of samples with fixed Co thickness of 6nm and various Ag thickness. One set of samples has the same total thickness 720nm and the other has the same bilayer number 60. For both sets, II seems to slowly increase as Ag gets thicker. From Tables 5.1, 5.2 we see that this occurs because the CIP MR becomes smaller faster than the CPP MR as adjacent Co layers get further apart. One way to explain this finding is that fewer electrons in the CIP geometry reach adjacent Co layers when the Ag spacer layers get thicker, while all the conduction electrons in the

| t <sub>Ag</sub> | M  | sample | C     | PP MI | 2%        | C     | IP MF | <b>R%</b> |      |
|-----------------|----|--------|-------|-------|-----------|-------|-------|-----------|------|
|                 |    | no.    | $H_0$ | $H_p$ | error     | $H_0$ | $H_p$ | error     |      |
| 2               | 90 | 251-01 | 1.1   | 1.3   | ±0.4      | 0.4   | 0.3   | ±0.1      |      |
| 4               | 72 | 251-03 | 73.9  | 37.0  | $\pm 0.5$ | 8.9   | 6.1   | ±0.1      | badª |
| 6               | 60 | 298-01 | 66.0  | 32.4  | ±0.8      | 19.0  | 8.7   | ±0.1      |      |
| 6               | 60 | 310-01 | 85.6  | 41.2  | ±0.3      | 7.3   | 4.4   | ±0.1      | bad  |
| 6               | 60 | 339-01 | 81.0  | 39.0  | ±0.2      | 25.8  | 11.5  | ±0.1      |      |
| 9               | 48 | 251-04 | 68.5  | 45.3  | ±1.1      | 12.8  | 8.8   | ±0.1      | bad  |
| 12              | 40 | 251-05 | 58.3  | 43.2  | ±1.2      | 12.3  | 8.4   | ±0.1      | bad  |
| 12              | 40 | 252-01 | 69.3  | 48.0  | ±1.2      | 19.1  | 12.0  | ±0.1      |      |
| 15              | 34 | 256-01 | 57.5  | 38.3  | ±1.2      | 6.4   | 4.0   | ±0.1      | bad  |
| 18              | 30 | 252-02 | 59.2  | 40.2  | $\pm 1.5$ | 8.4   | 5.6   | ±0.1      | bad  |
| 22              | 26 | 252-03 | 50.0  | 33.3  | ±1.4      | 9.3   | 5.9   | ±0.1      | bad  |
| 30              | 20 | 336-02 | 41.6  | 27.5  | $\pm 0.5$ | 7.5   | 5.2   | ±0.1      |      |
| 35              | 18 | 256-05 | 38.9  | 22.9  | ±1.9      | 6.3   | 4.4   | ±0.1      |      |
| 60              | 11 | 256-02 | 20.4  | 11.4  | ±1.9      | 1.5   | 0.9   | ±0.1      | bad  |
| 60              | 11 | 338-02 | 21.4  | 13.0  | ±0.6      | 1.9   | 1.4   | ±0.1      | bad  |

Table 5.1. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(6nm)/Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.

Table 5.2. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(6nm)/Ag(t)] \times 60$  samples. The right side is data for simple square films.

| $t_{Ag}$ | sample | CI    | PP M  | R%        | C     | IP MI | R%    |      | sample | C     | IP MI | <b>R</b> % |
|----------|--------|-------|-------|-----------|-------|-------|-------|------|--------|-------|-------|------------|
|          | no.    | $H_0$ | $H_p$ | error     | $H_0$ | $H_p$ | error |      | no.    | $H_0$ | $H_p$ | error      |
| 4        | 300-01 | 33.9  | 16.3  | ±0.6      | 6.6   | 4.1   | ±0.1  |      | 300-06 | 18.5  | 10.0  | ±0.5       |
| 4        | 338-05 | 75.4  | 35.0  | ±0.6      | 20.0  | 10.7  | ±0.1  |      |        |       |       |            |
| 4        | 350-04 | 49.3  | 23.8  | ±0.3      | 2.9   | 2.1   | ±0.1  | badª |        |       |       |            |
| 5        | 310-03 | 82.6  | 40.6  | ±0.5      | 22.0  | 11.6  | ±0.1  |      | 310-07 | 12.8  | 6.5   | ±0.5       |
| 6        | 298-01 | 66.0  | 32.4  | ±0.8      | 19.0  | 8.7   | ±0.1  |      | 298-06 | 14.1  | 11.5  | $\pm 0.5$  |
| 6        | 310-01 | 85.6  | 41.2  | ±0.3      | 7.3   | 4.4   | ±0.1  | bad  | 310-06 | 19.9  | 11.6  | ±0.5       |
| 6        | 339-01 | 81.0  | 39.0  | ±0.2      | 25.8  | 11.5  | ±0.1  |      | 308-07 | 19.4  | 10.5  | ±0.5       |
| 9        | 310-04 | 77.0  | 48.7  | ±0.4      | 20.4  | 13.6  | ±0.1  |      | 310-08 | 12.9  | 10.4  | ±0.5       |
| 12       | 300-03 | 71.4  | 47.1  | $\pm 1.5$ | 20.3  | 13.0  | ±0.1  |      | 300-07 | 12.4  | 8.7   | ±0.5       |
| 12       | 338-07 | 70.7  | 45.9  | ±0.3      | 8.4   | 5.6   | ±0.1  | bad  |        |       |       |            |
| 15       | 308-02 | 76.4  | 53.0  | ±0.2      | 17.0  | 13.1  | ±0.1  |      | 308-06 | 10.5  | 7.8   | ±0.5       |
| 15       | 350-07 | 75.5  | 51.9  | ±0.3      | 9.1   | 6.1   | ±0.1  | bad  |        |       |       |            |
| 20       | 338-06 | 63.1  | 36.9  | ±0.4      | 13.8  | 9.8   | ±0.1  |      |        |       |       |            |
| 30       | 350-08 | 62.2  | 40.7  | ±0.3      | 11.7  | 8.5   | ±0.1  |      |        |       |       |            |

<sup>a</sup>Misaligned CIP sample (see section 2.2).

Table 5.3. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(t)/Ag(6nm)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t <sub>Co</sub> | M  | sample | $\begin{array}{c c} CPP MR\% \\ \hline H_0 & H_p & err \\ \hline 100.0 & 65.9 & \pm 1 \\ \hline 99.8 & 64.2 & \pm 0 \\ \hline 104.9 & 66.8 & \pm 0 \\ \hline 104.2 & 67.1 & \pm 0 \\ \hline 98.1 & 61.0 & \pm 1 \\ \hline 66.0 & 32.4 & \pm 0 \\ \hline 85.6 & 41.2 & \pm 0 \\ \hline 81.0 & 39.0 & \pm 0 \\ \hline 66.4 & 20.0 & \pm 0 \\ \hline \end{array}$ |              |       | C     | IP MI | R%    |      | sample | C     | IP MI | R%        |
|-----------------|----|--------|--|--------------|-------|-------|-------|-------|------|--------|-------|-------|-----------|
|                 |    | no.    | H <sub>0</sub>   | $H_p$        | error | $H_0$ | $H_p$ | error |      | no.    | $H_0$ | $H_p$ | error     |
| 1.5             | 96 | 276-01 | 100.0  | 65.9         | ±1.0  | 29.4  | 22.2  | ±0.1  |      | 276-06 | 25.2  | 21.5  | ±0.5      |
| 2               | 90 | 261-01 | 99.8   | 64.2         | ±0.5  | 31.9  | 21.9  | ±0.1  |      | 261-06 | 24.4  | 21.0  | ±0.5      |
| 2               | 90 | 338-04 | 104.9  | <b>66.</b> 8 | ±0.6  | 31.7  | 21.3  | ±0.1  |      |        |       |       |           |
| 2               | 90 | 340-01 | 104.2  | 67.1         | ±0.6  | 32.6  | 19.6  | ±0.1  |      |        |       |       |           |
| 4               | 72 | 276-02 | 98.1   | 61.0         | ±1.1  | 28.6  | 17.9  | ±0.1  |      | 276-07 | 10.7  | 14.4  | ±0.5      |
| 6               | 60 | 298-01 | 66.0   | 32.4         | ±0.8  | 19.0  | 8.7   | ±0.1  |      | 298-06 | 14.1  | 11.5  | $\pm 0.5$ |
| 6               | 60 | 310-01 | 85.6   | 41.2         | ±0.3  | 7.3   | 4.4   | ±0.1  | badª | 310-06 | 19.9  | 11.6  | ±0.5      |
| 6               | 60 | 339-01 | 81.0   | 39.0         | ±0.2  | 25.8  | 11.5  | ±0.1  |      | 308-07 | 19.4  | 10.5  | ±0.5      |
| 9               | 48 | 339-02 | 66.4   | 20.0         | ±0.5  | 9.7   | 4.0   | ±0.1  |      |        |       |       |           |
| 12              | 40 | 276-03 | 53.6   | 11.1         | ±1.0  | 17.2  | 6.6   | ±0.1  |      | 276-08 | 6.7   | 2.4   | $\pm 0.5$ |
| 12              | 40 | 338-03 | 33.7   | 10.4         | ±0.4  | 11.9  | 4.4   | ±0.1  |      | 261-11 | 4.5   | 5.4   | $\pm 0.5$ |
| 30              | 20 | 340-02 | 2.2  | 2.0          | ±0.3  |       |       |       | б    |        |       |       |           |
| 60              | 11 | 276-04 | 4.1  | 5.6          | ±1.0  |       |       |       | Б    | 276-09 | 0.0   | 0.0   | ±0.5      |

<sup>b</sup>CIP resistance of this sample is not a smooth function of magnetic field, no peak and saturation values can be defined.

Table 5.4. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(t)/Ag(6nm)] \times 60$  samples. The right side is data for simple square films.

| t <sub>Co</sub> | sample | CI    | PP M  | R%        | C     | IP MI | R%    |      | sample | C     | IP MI | <b>R</b> % |
|-----------------|--------|-------|-------|-----------|-------|-------|-------|------|--------|-------|-------|------------|
|                 | no.    | $H_0$ | $H_p$ | error     | $H_0$ | $H_p$ | error |      | no.    | $H_0$ | $H_p$ | error      |
| 2               | 297-01 | 87.9  | 57.0  | $\pm 2.1$ | 27.5  | 20.4  | ±0.1  |      | 297-06 | 24.0  | 16.6  | $\pm 0.5$  |
| 6               | 298-01 | 66.0  | 32.4  | ±0.8      | 19.0  | 8.7   | ±0.1  |      | 298-06 | 14.1  | 11.5  | ±0.5       |
| 6               | 310-01 | 85.6  | 41.2  | ±0.3      | 7.3   | 4.4   | ±0.1  | badª | 310-06 | 19.9  | 11.6  | $\pm 0.5$  |
| 6               | 339-01 | 81.0  | 39.0  | ±0.2      | 25.8  | 11.5  | ±0.1  |      | 308-07 | 19.4  | 10.5  | ±0.5       |
| 9               | 297-03 | 62.2  | 25.4  | ±0.9      | 20.4  | 8.3   | ±0.1  |      | 297-08 | 1.8   | 1.2   | ±0.5       |
| 12              | 352-01 | 36.5  | 10.3  | ±0.2      | 3.9   | 1.8   | ±0.1  | bad  |        |       |       |            |
| 15              | 308-01 | 25.3  | 10.1  | ±0.8      | 5.0   | 3.6   | ±0.1  |      | 308-10 | 8.7   | 3.0   | ±0.5       |
| 18              | 352-02 | 6.0   | 4.0   | ±0.2      | 1.1   | 0.9   | ±0.1  | bad  |        |       |       |            |
| 20              | 298-03 | 3.8   | 3.3   | ±0.4      | 1.2   | 1.3   | ±0.1  |      | 298-08 | 1.7   | 1.1   | ±0.5       |

<sup>a</sup>Misaligned CIP sample (see section 2.2).

Table 5.5. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(t)=Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t  | M  | sample | CI    | PP M  | R%        | C     | IP MI | R%    |                    | sample | CI               | P MR  | 2%        |
|----|----|--------|-------|-------|-----------|-------|-------|-------|--------------------|--------|------------------|-------|-----------|
|    |    | no.    | $H_0$ | $H_p$ | error     | $H_0$ | $H_p$ | error |                    | no.    | $H_0$            | $H_p$ | error     |
| 6  | 60 | 298-01 | 66.0  | 32.4  | ±0.8      | 19.0  | 8.7   | ±0.1  |                    | 298-06 | 14.1             | 11.5  | $\pm 0.5$ |
| 6  | 60 | 310-01 | 85.6  | 41.2  | ±0.3      | 7.3   | 4.4   | ±0.1  | $\mathbf{bad}^{a}$ | 310-06 | 19.9             | 11.6  | $\pm 0.5$ |
| 6  | 60 | 339-01 | 81.0  | 39.0  | ±0.2      | 25.8  | 11.5  | ±0.1  |                    | 308-07 | 19.4             | 10.5  | $\pm 0.5$ |
| 8  | 45 | 246-05 | 65.3  | 36.8  | ±1.1      | 15.4  | 9.2   | ±0.1  |                    | 246-08 | N/A <sup>b</sup> | 8.7   | $\pm 0.5$ |
| 12 | 30 | 337-02 | 54.6  | 27.8  | ±0.3      |       |       |       |                    |        |                  |       |           |
| 15 | 24 | 246-09 | 39.4  | 22.2  | $\pm 1.2$ | 9.7   | 5.8   | ±0.1  |                    | 246-11 | N/A <sup>b</sup> | 3.2   | $\pm 0.5$ |
| 18 | 20 | 336-03 | 37.4  | 17.6  | $\pm 0.3$ | 7.4   | 4.0   | ±0.1  |                    |        |                  |       |           |
| 20 | 18 | 260-01 | 33.1  | 13.0  | ±0.6      | 4.3   | 2.2   | ±0.1  | bad                | 260-09 | 1.8              | 1.8   | $\pm 0.5$ |
| 30 | 12 | 260-02 | 21.7  | 7.0   | ±0.9      | 2.6   | 1.5   | ±0.1  | bad                | 260-10 | 0.0              | 0.0   | ±0.8      |
| 50 | 7  | 260-03 | 11.1  | 6.7   | $\pm 1.3$ | 0.9   | 0.5   | ±0.1  | bad                | 260-11 | 0.0              | 0.0   | ±0.5      |

<sup>b</sup>We didn't appreciate the irreversibility of the zero field value for the first few samples, which were saturated before any measurements were made.

Table 5.6. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(6nm)/Ag(6nm)] \times M$  samples. 0.5 bilayer means a Co layer.

| M    | sample | C     | PP MI | <b>R%</b> | C     |       |       |                  |
|------|--------|-------|-------|-----------|-------|-------|-------|------------------|
|      | no.    | $H_0$ | $H_p$ | error     | $H_0$ | $H_p$ | error |                  |
| 1.5  | 337-03 | 4.1   | 3.2   | $\pm 2.8$ |       |       |       | a                |
| 3.5  | 337-04 | 17.0  | 11.1  | $\pm 5.1$ | 3.9   | 3.4   | ±0.1  |                  |
| 5.5  | 337-05 | 26.7  | 16.5  | $\pm 2.1$ | 6.9   | 3.5   | ±0.1  |                  |
| 10.5 | 350-01 | 41.0  | 21.6  | $\pm 1.3$ | 5.3   | 3.3   | ±0.1  | bad <sup>6</sup> |
| 20.5 | 350-02 | 60.5  | 32.8  | ±0.9      | 13.4  | 7.2   | ±0.2  |                  |
| 30.5 | 350-03 | 73.4  | 33.8  | ±0.7      | 8.9   | 4.2   | ±0.1  | bad              |
| 60   | 298-01 | 66.0  | 32.4  | ±0.8      | 19.0  | 8.7   | ±0.1  |                  |
| 60   | 310-01 | 85.6  | 41.2  | $\pm 0.3$ | 7.3   | 4.4   | ±0.1  | bad              |
| 60   | 339-01 | 81.0  | 39.0  | ±0.2      | 25.8  | 11.5  | ±0.1  |                  |

<sup>a</sup>CIP resistance of this sample is not a smooth function of magnetic field, no peak and saturation values can be defined.

<sup>b</sup>Misaligned CIP sample (see section 2.2).

Table 5.7. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(2nm)/Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t <sub>Ag</sub> | M  | sample | CP    | P MI  | <b>R</b> % | CI    | CIP MR% |       |         | sample | C     | IP MI | <b>R%</b> |
|-----------------|----|--------|-------|-------|------------|-------|---------|-------|---------|--------|-------|-------|-----------|
|                 |    | no.    | $H_0$ | $H_p$ | error      | $H_0$ | $H_p$   | error |         | no.    | $H_0$ | $H_p$ | error     |
| 6               | 90 | 261-01 | 99.8  | 64.2  | ±0.5       | 31.9  | 21.9    | ±0.1  |         | 261-06 | 24.4  | 21.0  | ±0.5      |
| 6               | 90 | 338-04 | 104.9 | 66.8  | ±0.6       | 31.7  | 21.3    | ±0.1  |         |        |       |       |           |
| 6               | 90 | 340-01 | 104.2 | 67.1  | ±0.6       | 32.6  | 19.6    | ±0.1  |         |        |       |       |           |
| 8               | 72 | 261-02 | 93.4  | 61.3  | ±1.2       | 12.3  | 8.1     | ±0.1  | $bad^a$ | 261-07 | 24.5  | 17.0  | ±0.5      |
| 12              | 51 | 261-03 | 77.2  | 45.3  | ±1.4       | 7.0   | 4.5     | ±0.1  | bad     | 261-08 | 9.2   | 11.8  | $\pm 0.5$ |
| 20              | 33 | 261-04 | 57.5  | 32.2  | $\pm 3.6$  | 13.0  | 8.3     | ±0.1  |         | 261-10 | 4.6   | 6.6   | $\pm 0.5$ |
| 58              | 12 | 261-05 | 19.8  | 11.1  | $\pm 1.1$  | N/A   | 2.2     | ±0.1  |         | 261-09 | 3.4   | 2.1   | ±0.5      |

Table 5.8. CPP and CIP MR values at  $H_0$  and  $H_p$  of  $[Co(6nm)/AgSn(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t <sub>AgSn</sub> | M  | sample | CPP MR% |       |           | C     | IP M  | IR%   | sample | C     | IP M  | IR%       |   |
|-------------------|----|--------|---------|-------|-----------|-------|-------|-------|--------|-------|-------|-----------|---|
|                   |    | no.    | $H_0$   | $H_p$ | error     | $H_0$ | $H_p$ | error | no.    | $H_0$ | $H_p$ | error     |   |
| 2                 | 90 | 262-01 | 0.8     | 0.9   | ±0.9      | 0.2   | 0.3   | ±0.1  | 262-06 |       |       |           | a |
| 4                 | 72 | 262-02 | 29.5    | 13.2  | $\pm 1.2$ | 3.9   | 1.9   | ±0.1  | 262-10 | 1.4   | 1.9   | ±0.5      |   |
| 6                 | 60 | 262-03 | 25.0    | 13.6  | ±0.9      | 3.4   | 1.7   | ±0.1  | 262-08 | 0.9   | 0.8   | ±0.5      |   |
| 6                 | 60 | 277-04 | 28.4    | 14.5  | $\pm 0.5$ | 4.1   | 2.1   | ±0.1  | 277-06 | 0.4   | 1.1   | $\pm 0.5$ |   |
| 12                | 40 | 277-02 | 11.7    | 6.9   | ±0.4      | 0.9   | 0.5   | ±0.1  | 277-07 |       |       |           | a |
| 15                | 34 | 262-04 | 7.7     | 4.7   | ±0.3      | 0.5   | 0.3   | ±0.1  | 262-07 |       |       |           | a |
| 35                | 18 | 262-05 | 2.0     | 1.1   | ±0.3      | 0.0   | 0.0   | ±0.1  | 262-09 |       |       |           | a |
| 60                | 11 | 277-03 | 0.8     | 0.3   | ±0.3      | 0.0   | 0.0   | ±0.1  | 277-08 |       |       |           | a |

<sup>a</sup>CIP resistance of this sample is not a smooth function of magnetic field, no peak and saturation values can be defined.



Figure 5.1. The ratio between CPP MR and CIP MR,  $\Pi$ , at  $H_0$  and  $H_p$  versus Ag thickness for two sets of samples with fixed Co thickness:  $[Co(6nm)/Ag(t)] \times M$  with total sample thickness 720nm, and  $[Co(6nm)/Ag(t)] \times 60$ .



Figure 5.2. The ratio between CPP MR and CIP MR, II, at  $H_0$  and  $H_p$  versus Co thickness for two sets of samples with fixed Ag thickness:  $[Co(t)/Ag(6nm)] \times M$  with total sample thickness 720nm, and  $[Co(t)/Ag(6nm)] \times 60$ .

CPP geometry go through the whole sample. Thus the thicker Ag layers isolate Co layers in CIP but not in CPP. This explanation agrees with theory, which finds that the elastic mean free paths are fundamental lengths in the CIP geometry but not in the CPP geometry (see e.g. [64, 147]).

Figure 5.2 shows a similar graph for samples with fixed Ag at 6nm and various Co thicknesses. From Tables 5.3, 5.4 we see that both the CPP and CIP MRs decrease as the Co thickness gets larger. However, with one exception,  $\Pi$  seems to be nearly constant. This behavior suggests that increasing Co thickness affects both the CPP and CIP MR in a similar way.

When the Co and Ag layers get thicker at the same time, as in the set of samples listed in Table 5.5,  $\Pi$  may get larger. Unfortunately we have too many misaligned CIP samples in this set of samples to reach any clear conclusion.

For the one set of Co/AgSn samples,  $\Pi$  is about 7 for samples with AgSn thickness 6nm or smaller and gets larger than 10 when AgSn layer is thicker than 12nm. From Table 5.8 we see that both the larger values of  $\Pi$  and the more rapid increase with spacer layer thicknesses are due to smaller CIP MR resulting from the much larger resistivity of the AgSn alloy.

## 5.1.3 Magnetic Ordering

### **Changing Ag Thickness**

To find out whether there is magnetic coupling across the spacer layers, and to investigate the magnetic ordering within and between the Co layers in our samples, we focus on samples having fixed Co thickness, 6nm, and varying Ag thicknesses. We also compare with Co/Cu multilayers, which is a project carried out by P. Holody and P.A. Schroeder at MSU. Tables 5.1 and 5.2 list two sets of Co/Ag samples, the former has fixed total sample thickness 720nm and the latter has the same bilayer number 60. Figure 5.3 shows the zero field CPP MR of these two sets of samples versus Ag thickness. We see that the sample with 2nm Ag has a MR close to zero, samples with 4nm Ag have MRs ranging from about 35% to 75%, samples with 6nm Ag have MRs ranging from about 66% to 86% and their average value is the largest in the set. Samples with Ag layers thicker than 6nm have decreasing MR with increasing Ag thicknesses.

Figure 5.5 shows the MR at 4K versus Cu thicknesses of a set of Co/Cu samples with Fe buffer layers and fixed Co thickness [162]. Both the CPP and CIP MR show clear oscillations with the amplitude diminishing with increasing Cu thickness. The large MR peak represents strong antiferromagnetic ordering in the sample at  $H_0$  and the zero MR represents strong ferromagnetic ordering at  $H_0$ . Both these orderings weaken as the Cu thickness increases, and the oscillations stop around Cu thickness 5-6nm.

For Co/Ag, no oscillations in MR have been reported for Co layers as thick as we used. Our data suggest that when the Ag layer is 4nm or thinner, there is ferromagnetic ordering in the multilayer samples. The small MR when the Ag layer is 2nm shows that the ordering is strong. The scatter of the MR data at 4nm Ag shows that the ordering is weaker across 4nm Ag and probably depends on detailed sample characteristics, like individual layer thickness variations. For samples with 6nm Ag, we presume that any ferromagnetic ordering at  $H_0$  is weak because of the Co/Cu multilayer data mentioned above and the consistent CPP AR behavior presented in section 3.2.3. The scatter of the data at 6nm Ag is simply taken as the sample reproducibility.

Figure 5.4 shows a graph for the CPP MR at  $H_p$  of the same two sets of samples as in Figure 5.3. Most of the features we described in CPP MR at  $H_0$  still apply to the MR at  $H_p$ , except that samples with 6nm Ag do not have the largest peak MR. It is clear from the graph that peak MRs of 6nm Ag samples are smaller than those



Figure 5.3. CPP MR at  $H_0$  versus Ag thickness for samples (a)  $[Co(6nm)/Ag(t)] \times M$ , total sample thickness 720nm, and (b)  $[Co(6nm)/Ag(t)] \times 60$ .



Figure 5.4. CPP MR at  $H_p$  versus Ag thickness for samples (a)  $[Co(6nm)/Ag(t)] \times M$ , total sample thickness 720nm, and (b)  $[Co(6nm)/Ag(t)] \times 60$ .



Figure 5.5. CPP and CIP MR versus Cu thickness for samples  $Fe(5nm)/[Co(6nm)/Cu(t)] \times M$ , total sample thickness 360nm. The lines are guides to the eye. (From reference [162].)

of 9nm and 12nm samples. To understand this difference from the MR behavior at  $H_0$ , we need to study sample structure in detail, but this study is not yet available.

CIP MR data at  $H_0$  and  $H_p$  for these two sets of samples show the same qualitative features we described above (see Tables 5.1 and 5.2).

### 5.1.4 Significant Interface Contribution to MR

We focus on Table 5.5 in this section. These samples have Ag layers at least 6nm thick, where as described above we expect weak or no ferromagnetic ordering in our samples. From these samples we hope to infer the importance of interface effects on the CPP MR.

In Table 5.5 we list a set of samples with equal amount of bulk materials, i.e. 360nm thick of Co and of Ag in each sample. By changing the bilayer thickness, we have different numbers of interfaces in each sample. As the number of the interfaces grows, the MR increases. This suggests that the interfacial effect on MR is significant. Quantitative analysis in section 5.5 verifies this conclusion.

### 5.1.5 MR versus Increasing Bilayers

Table 5.6 is a set of samples with fixed Co and Ag thicknesses at 6nm and increasing bilayer number. The MR is small for samples with only a few bilayers, increases rapidly with increasing bilayer number, and levels off at about 20 bilayers. This behavior can be understood by a simple argument: when the bilayer number is increased, one adds to equation 5.1 the quantity a in the numerator and the quantity b in the denominator for each bilayer. Eventually the MR ratio will become a/b.

The Nb contact resistance is especially important in the CPP MR for those samples in the set with only a few bilayers. These samples have very small intrinsic resistances and the Nb contact resistance reduces the CPP MR significantly. However, the above qualitative analysis remains true, even after subtracting the Nb contact resistance.

## 5.1.6 MR( $H_0$ ) versus MR( $H_p$ )

Almost all of our samples have their largest resistances (or MRs) at  $H_0$  before any magnetic field is applied. Once a magnetic field is applied, the resistances at  $H_0$  (see e.g. Figure 3.2) do not usually seem to be retrievable. Efforts were made to bring the resistance back up to the  $H_0$  value, but these were successful in only one case. We warmed samples up to room temperature and cooled them down again to 4K; the resistances did not exceed the peak values. By decreasing the field just after the sample resistance passed its peak, we did see the resistance increase beyond the peak value for samples with Co thickness 6nm or smaller. When the field direction was reversed, the resistance went through a maximum at a field smaller than  $H_p$ . Applying this procedure several times, we reached a point where the resistance had a 'final maximum' value around zero applied field; increasing or decreasing the field both reduced the resistance. For those few samples we tried this procedure on, 6nm thick Co samples showed 'final maximum' resistance just a bit larger than the  $H_p$ value defined in our usual way, but still much smaller than the  $H_0$  value. For samples with 1.5nm Co layers, we were able to increase the resistances to values half way between  $H_0$  and  $H_p$  values or more. One such sample had a 'final maximum' slightly larger than the  $H_0$  value.

There are at least two different ways to model the magnetic structures in our samples.

In one model, the Co layers in our as-deposited sample are taken to be large single domains aligned anti-ferromagnetically from one layer to the next. Once an external magnetic field is applied, these single domains break into smaller ones to align their moments with the field. The sample cannot return to the large single domain structure by temperature and magnetic field variations. The as-deposited sample has a magnetic structure closest to the AF coupled state assumed in theories, and thus the largest resistance.

In the other model, the as-deposited sample is taken to have very small domains, and thus many domain boundaries. These boundaries cause extra resistance, and they are eliminated by applying an external field. When the external magnetic field is reversed, Co layers break into smaller domains but not as small as the domains in the as-deposited samples. Thus the sample resistance can never get back to its as-deposited value. Detailed structure characterizations are needed to resolve this issue. Such characterizations are not yet available.

Trying to find some clue for the structures of our samples at  $H_0$  and  $H_p$ , we look at the ratio between MRs at  $H_0$  and  $H_p$ . Figure 5.6 is a graph of this ratio in both CPP and CIP geometries. We plot two sets of samples with fixed Co at 6nm and one set with equal Co and Ag thicknesses. We see that the ratios are similar for Ag thicknesses 4nm and larger. The ratio of about unity for the sample with 2nm Ag is consistent with the ferromagnetic ordering inferred above. Samples with equal Co and Ag thicknesses have consistently higher ratios than the other two sets with fixed Co thickness (6nm). The significance of this is not yet understood.

Figure 5.7 is a similar graph versus Co thickness for two sets of samples with fixed 6nm Ag and variable Co thickness. The ratio of the MRs at  $H_0$  and  $H_p$  increases quickly with Co thickness, reaches a maximum at 12nm, and then fall to about unity. This tells us that, when the Co thickness increases beyond 12nm, the MR at  $H_0$  decreases faster than the MR at  $H_p$  does. At 20nm, the MR at  $H_0$  becomes comparable to the MR at  $H_p$  and the ratio is close to one. Thus our data suggest that, when the Ag thickness is fixed at 6nm, there is structural change within each individual Co layer as the Co thickness increases beyond 12nm.

The remaining three sets of samples, all with fixed Co thickness, have ratios of MR at  $H_0$  and  $H_p$  ranging between 1.2 and 2.5.

To summarize, our data suggest that the magnetic structures of our samples are much alike when the Co thickness is kept fixed. When the Co thickness is varied, the crystalline structure and/or magnetic structure of the Co layer can change. These different structures cause the ratio of MRs at  $H_0$  and  $H_p$  to change when the Co thickness changes.



Figure 5.6. The ratio between MRs at  $H_0$  and  $H_p$ ,  $\pi$ , of CPP and CIP geometry versus Ag thickness for two sets of samples with fixed Co thickness:  $[Co(6nm)/Ag(t)] \times M$  with total sample thickness 720nm, and  $[Co(6nm)/Ag(t)] \times 60$ ; and one set of samples having equal thickness for Co and Ag:  $[Co(t)=Ag(t)] \times M$  with total sample thickness 720nm.



Figure 5.7. The ratio between the MRs at  $H_0$  and  $H_p$ ,  $\pi$ , of CPP and CIP geometry versus Co thickness for two sets of samples with fixed Ag thickness:  $[Co(t)/Ag(6nm)] \times M$  with total sample thickness 720nm, and  $[Co(t)/Ag(6nm)] \times 60$ .
#### 5.2 Magnetic Moments versus Field

Magnetic moment measurements on our multilayer samples were made near 4K. The magnetic moment  $\mathcal{M}$  versus field curves for most of our multilayer samples have similar shape. Samples with 20nm thick Co layer and larger sometimes show shoulders and steps on the curves, with all the features reversible, i.e. if we cycle the field or warm the sample to room temperature and then cool down near 4K, the shoulders and steps stay the same. In Figure 5.8 we show some of the  $\mathcal{M}$  versus field curves of our samples, in addition to Figures 3.2 and 3.3 already shown in chapter 3. Parts (a) and (b) are samples in our perpendicular sample shape grown on sapphire. They were measured at 15K and 12K, respectively, so that the Nb is not superconducting. All other samples in the figure are simple multilayers on Si, measured at 5K.

We checked for systematic changes in remanence  $\mathcal{M}_r$  and coercive field  $H_c$  in our samples. Upon reducing the applied field from saturation, the remanence is defined as the residual magnetic moment when the external field is zero, and the coercive field is defined as the reverse field needed to reduce the moment to zero. The coercive fields are listed in the tables in the following section, where they are compared with the peak fields of MR. The remanence of most our multilayers ranges from  $0.8\mathcal{M}_{\bullet}$  to  $0.9\mathcal{M}_{\bullet}$ , as in Figure 5.8 (b) and (c), with no apparent systematic dependence on Co or Ag thicknesses. Four samples had smaller remanence; they are shown in Figure 5.8 (a), (d), (e), and (f). The first three had remanences of  $0.60\mathcal{M}_{\bullet}$ ,  $0.63\mathcal{M}_{\bullet}$ ,  $0.53\mathcal{M}_{\bullet}$ , respectively. The fourth, a  $[Co(6nm)/AgSn(2nm)] \times 90$  sample, had an asymmetrical curve with  $0.50\mathcal{M}_{\bullet}$  on one side and  $0.76\mathcal{M}_{\bullet}$  on the other.

Since shoulders and steps on magnetization curve are only seen in thick Co samples, we conclude that some magnetic sub-structure starts to develop when the Co layer is 20nm or thicker. We don't have a good explanation for the smaller remanence seen on some samples. More systematic studies on single Co thin films and Co/Ag/Co



Figure 5.8. Magnetic moment versus field for six multilayer samples.

sandwiches are needed to understand the behavior of our multilayers.

## 5.3 CPP and CIP MR Peak Fields and Coersive Fields

The peak magnetic fields,  $H_p$ , are listed in this section together with the coercive fields,  $H_c$ . The peak field is where the maxima of the CPP and CIP MRs occur after the multilayers have been taken to above saturation, and the applied field is reversed.

The values of  $H_p$  for CIP listed next to those for CPP are calculated according to the calibration for the CPP sample. Since the effective sample for CIP is at a different position in the magnet than that for the CPP, the listing has systematic bias. Misaligned CIP samples sometimes have much smaller  $H_ps$ , they are listed in the tables with smaller font.

#### 5.3.1 Experimental Results

See Table 5.9 to Table 5.16.

From the tables, we see that the  $H_p$ s for CPP and for CIP geometries are comparable.  $H_c$  is smaller than both the  $H_p$ s for most samples.

When Co layer thickness is kept fixed, all three fields increase with increasing Ag thickness. When Ag layer thickness is kept fixed, all fields decrease first then increase with Co thickness. When Co and Ag thicknesses are kept equal and changing, all fields are roughly independent of layer thickness until the Co layer is thicker than 20nm; then we see structure in resistance and magnetization curves. We do not have good interpretations for these experimental facts so far; detailed sample structure studies would reveal the underlying physics.

| $t_{Ag}$ | M  | sample |     | $H_p(Oe)$ | )        |      |
|----------|----|--------|-----|-----------|----------|------|
|          |    | no.    | CPP | CIP       | error    |      |
| 2        | 90 | 251-01 | 65  | 22        | $\pm 15$ |      |
| 4        | 72 | 251-03 | 103 | 86        | $\pm 15$ | badª |
| 6        | 60 | 298-01 | 103 | 91        | $\pm 15$ |      |
| 6        | 60 | 310-01 | 74  | 69        | $\pm 15$ | bad  |
| 9        | 48 | 251-04 | 99  | 91        | $\pm 15$ | bad  |
| 12       | 40 | 251-05 | 120 | 112       | $\pm 15$ | bad  |
| 12       | 40 | 252-01 | 112 | 95        | $\pm 15$ |      |
| 15       | 34 | 256-01 | 108 | 99        | $\pm 15$ | bad  |
| 18       | 30 | 252-02 | 129 | 120       | $\pm 15$ | bad  |
| 22       | 26 | 252-03 | 142 | 129       | $\pm 15$ | bad  |
| 30       | 20 | 336-02 | 125 | 125       | $\pm 15$ |      |
| 35       | 18 | 256-05 | 154 | 137       | $\pm 15$ |      |
| 60       | 11 | 256-02 | 200 | 171       | $\pm 15$ | bad  |
| 60       | 11 | 338-02 | 150 | 150       | $\pm 15$ | bad  |

Table 5.9. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(6nm)/Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.

Table 5.10. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(6nm)/Ag(t)] \times 60$  samples. The right side is data for simple square films.

| $t_{Ag}$ | sample |     | $H_p(Oe$ | )        |      | sample | $H_p(Oe)$ | $H_c(Oe)$ |       |
|----------|--------|-----|----------|----------|------|--------|-----------|-----------|-------|
|          | no.    | CPP | CIP      | error    |      | no.    | CIP       |           | error |
| 4        | 300-01 | 78  | 65       | $\pm 15$ |      | 300-06 | 80        | 55        | ±10   |
| 4        | 338-05 | 99  | 86       | $\pm 15$ |      |        |           |           |       |
| 4        | 350-04 | 78  | 43       | $\pm 15$ | badª |        |           |           |       |
| 5        | 310-03 | 86  | 65       | $\pm 15$ |      | 310-07 | 80        | 60        | ±10   |
| 6        | 298-01 | 103 | 91       | $\pm 15$ |      | 298-06 | 90        | 75        | ±10   |
| 6        | 310-01 | 74  | 69       | $\pm 15$ | bad  | 310-06 | 80        | 65        | ±10   |
| 9        | 310-04 | 91  | 78       | $\pm 15$ |      | 310-08 | 90        | 90        | ±10   |
| 12       | 300-03 | 95  | 86       | $\pm 15$ |      | 300-07 | 100       | 95        | ±10   |
| 12       | 338-07 | 116 | 116      | $\pm 15$ | bad  |        |           |           |       |
| 15       | 308-02 | 120 | 103      | $\pm 15$ |      | 308-06 | 120       | 115       | ±10   |
| 15       | 350-07 | 154 | 142      | $\pm 15$ | bad  |        |           |           |       |
| 20       | 338-06 | 179 | 162      | $\pm 15$ |      |        |           |           |       |
| 30       | 350-08 | 187 | 179      | $\pm 15$ |      |        |           |           |       |

<sup>a</sup>Misaligned sample (see section 2.2).

Table 5.11. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(t)/Ag(6nm)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t <sub>Co</sub> | M  | sample |     | $\overline{H_p(\text{Oe})}$ | )        |                  | sample | $H_p(Oe)$ | $H_c(Oe)$ |       |
|-----------------|----|--------|-----|-----------------------------|----------|------------------|--------|-----------|-----------|-------|
|                 |    | no.    | CPP | CIP                         | error    |                  | no.    | CIP       |           | error |
| 1.5             | 96 | 276-01 | 319 | 298                         | $\pm 15$ |                  | 276-06 | 440       | 375       | ±10   |
| 2               | 90 | 261-01 | 253 | 216                         | $\pm 15$ |                  | 261-06 | 230       | 190       | ±10   |
| 2               | 90 | 338-04 | 192 | 187                         | ±15      |                  |        |           |           |       |
| 2               | 90 | 340-01 | 212 | 171                         | $\pm 15$ |                  |        |           |           |       |
| 4               | 72 | 276-02 | 91  | 82                          | $\pm 15$ |                  | 276-07 | 90        | 90        | ±10   |
| 6               | 60 | 298-01 | 103 | 91                          | $\pm 15$ |                  | 298-06 | 90        | 75        | ±10   |
| 6               | 60 | 310-01 | 74  | 69                          | $\pm 15$ | bad <sup>a</sup> | 310-06 | 80        | 65        | ±10   |
| 9               | 48 | 339-02 | 69  | 69                          | ±15      | bad              |        |           |           |       |
| 12              | 40 | 276-03 | 78  | 69                          | $\pm 15$ | bad              | 276-08 | 100       | 75        | ±10   |
| 12              | 40 | 338-03 | 82  | 65                          | ±15      |                  | 261-11 | 60        | 55        | ±10   |
| 30              | 20 | 340-02 | 129 |                             | $\pm 15$ | 6                |        |           |           |       |
| 60              | 11 | 276-04 | 99  |                             | ±15      |                  | 276-09 |           | 90        | ±10   |

<sup>b</sup>CIP resistance of this sample is not a smooth function of magnetic field, no peak and saturation values can be defined.

Table 5.12. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(t)/Ag(6nm)] \times 60$  samples. The right side is data for simple square films.

| t <sub>Co</sub> | sample |     | $H_p(Oe$ | )        |         | sample | $H_p(Oe)$ | $H_c(Oe)$ |       |
|-----------------|--------|-----|----------|----------|---------|--------|-----------|-----------|-------|
|                 | no.    | CPP | CIP      | error    |         | no.    | CIP       |           | error |
| 2               | 297-01 | 229 | 187      | $\pm 15$ |         | 297-06 | 200       | 165       | ±10   |
| 6               | 298-01 | 103 | 91       | $\pm 15$ |         | 298-06 | 90        | 75        | ±10   |
| 6               | 310-01 | 74  | 69       | ±15      | $bad^a$ | 310-06 | 80        | 65        | ±10   |
| 9               | 297-03 | 78  | 69       | $\pm 15$ |         | 297-08 | 75        | 60        | ±10   |
| 12              | 352-01 | 69  | 61       | $\pm 15$ |         |        |           |           |       |
| 15              | 308-01 | 78  | 65       | ±15      |         | 308-10 | 80        | 85        | ±10   |
| 18              | 352-02 | 86  | 65       | $\pm 15$ |         |        |           |           |       |
| 20              | 298-03 | 129 | 120      | ±15      |         | 298-08 | 75        | 105       | ±10   |

<sup>a</sup>Misaligned sample (see section 2.2).



Table 5.13. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(t)=Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t  | M  | sample |     | $\overline{H_p(\text{Oe})}$ | )        |      | sample | $H_p(Oe)$ | $H_c(Oe)$ |       |
|----|----|--------|-----|-----------------------------|----------|------|--------|-----------|-----------|-------|
|    |    | no.    | CPP | CIP                         | error    |      | no.    | CIP       |           | error |
| 6  | 60 | 298-01 | 103 | 91                          | $\pm 15$ |      | 298-06 | 90        | 75        | ±10   |
| 6  | 60 | 310-01 | 74  | 69                          | ±15      | badª | 310-06 | 80        | 65        | ±10   |
| 8  | 45 | 246-05 | 78  | 78                          | $\pm 15$ |      | 246-08 | 80        | 65        | ±10   |
| 12 | 30 | 337-02 | 69  |                             | ±15      |      |        |           |           |       |
| 15 | 24 | 246-09 | 61  | 56                          | $\pm 15$ |      | 246-11 | 50        | 50        | ±10   |
| 18 | 20 | 336-03 | 43  | 56                          | ±15      |      |        |           |           |       |
| 20 | 18 | 260-01 | 61  | 43                          | $\pm 15$ | bad  | 260-09 | 45        | 45        | ±10   |
| 30 | 12 | 260-02 | 65  | 48                          | $\pm 15$ | bad  | 260-10 | 115       | 75        | ±     |
| 50 | 7  | 260-03 | 129 | 65                          | $\pm 15$ | bad  | 260-11 | 80        | 165       | ±10   |

Table 5.14. CPP and CIP  $H_p$  of  $[Co(6nm)/Ag(6nm)] \times M$  samples.

| M    | sample |     | $H_p(\text{Oe})$ |          |                  |  |  |
|------|--------|-----|------------------|----------|------------------|--|--|
|      | no.    | CPP | CIP              | error    |                  |  |  |
| 1.5  | 337-03 | 192 |                  | $\pm 15$ | a                |  |  |
| 3.5  | 337-04 | 82  | 78               | $\pm 15$ |                  |  |  |
| 5.5  | 337-05 | 99  | 43               | $\pm 15$ |                  |  |  |
| 10.5 | 350-01 | 69  | 48               | $\pm 15$ | bad <sup>b</sup> |  |  |
| 20.5 | 350-02 | 99  | 52               | $\pm 15$ |                  |  |  |
| 30.5 | 350-03 | 69  | 61               | ±20      | bad              |  |  |
| 60   | 298-01 | 103 | 91               | $\pm 15$ |                  |  |  |
| 60   | 310-01 | 74  | 69               | $\pm 15$ | bad              |  |  |

<sup>a</sup>CIP resistance of this sample is not a smooth function of magnetic field, no peak and saturation values can be defined.

<sup>b</sup>Misaligned sample (see section2.2).

Table 5.15. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(2nm)/Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| $t_{Ag}$ | M  | sample |     | $\overline{H_p(\text{Oe})}$ | )        |      | sample | $H_p(Oe)$ | $H_c(Oe)$ |          |
|----------|----|--------|-----|-----------------------------|----------|------|--------|-----------|-----------|----------|
|          |    | no.    | CPP | CIP                         | error    |      | no.    | CIP       |           | error    |
| 6        | 90 | 261-01 | 253 | 216                         | $\pm 15$ |      | 261-06 | 230       | 190       | ±10      |
| 6        | 90 | 338-04 | 192 | 189                         | $\pm 15$ |      |        |           |           |          |
| 6        | 90 | 340-01 | 212 | 171                         | ±15      |      |        |           |           |          |
| 8        | 72 | 261-02 | 274 | 274                         | $\pm 15$ | badª | 261-07 | 300       | 275       | ±10      |
| 12       | 51 | 261-03 | 375 | 351                         | $\pm 15$ | bad  | 261-08 | 400       | 360       | $\pm 50$ |
| 20       | 33 | 261-04 | 355 | 391                         | $\pm 15$ |      | 261-10 | 400       | 390       | ±10      |
| 58       | 12 | 261-05 | 375 | 396                         | $\pm 15$ |      | 261-09 | 430       | 385       | ±10      |

Table 5.16. CPP, CIP  $H_p$  and  $H_c$  of  $[Co(6nm)/AgSn(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm. The right side is data for simple square films.

| t <sub>AgSn</sub> | M  | sample |     | $H_p(Oe)$ | )        | sample | $H_p(Oe)$ | $H_c(Oe)$ |       |
|-------------------|----|--------|-----|-----------|----------|--------|-----------|-----------|-------|
|                   |    | no.    | CPP | CIP       | error    | no.    | CIP       |           | error |
| 2                 | 90 | 262-01 | 86  | 65        | $\pm 15$ | 262-06 |           | 20        | ±10   |
| 4                 | 72 | 262-02 | 108 | 82        | $\pm 15$ | 262-10 | 120       | 80        | ±10   |
| 6                 | 60 | 262-03 | 112 | 103       | $\pm 15$ | 262-08 | 120       | 100       | ±10   |
| 6                 | 60 | 277-04 | 120 | 95        | $\pm 15$ | 277-06 | 140       | 80        | ±10   |
| 12                | 40 | 277-02 | 133 | 129       | $\pm 15$ | 277-07 |           | 140       | ±10   |
| 15                | 34 | 262-04 | 112 | 95        | $\pm 15$ | 262-07 |           | 105       | ±10   |
| 35                | 18 | 262-05 | 129 | 86        | ±10      | 262-09 |           | 145       | ±10   |
| 60                | 11 | 277-03 | 175 |           | $\pm 15$ | 277-08 |           | 180       | ±10   |

#### 5.4 CPP Resistance

To quantitatively analyze our CPP AR data, we need to associate the data with a theoretical model. Obviously all Co layers have magnetization aligned along the external field at  $H_s$ , thus we associate our data at  $H_s$  with the ferromagnetic (F) arrangement of equation (4.10) in the two channel model. The magnetic configurations of our samples at  $H_0$  and  $H_p$  are poorly understood. Theoretically, single domain magnetic multilayers have their largest resistances when the magnetic moments of the magnetic layers are in an antiferromagnetic (AF) arrangement. We thus associate the largest CPP resistance, at  $H_0$ , with equation (4.9) in the two channel model. In appendix E, we show that our data at  $H_p$  show qualitatively similar behavior to the data at  $H_0$ . Thus we use the same equation, (4.9), to analyze the data at  $H_p$  there.

To reiterate, we assume that all the Co layers of our samples are in the spontaneously AF configuration at  $H_0$ , and in F configuration at  $H_s$ . Zhang and Levy [67] argued that the CPP resistance of a multilayer which has a magnetic structure consisting of a superposition of statistically un-correlated magnetic configurations with zero total magnetic moment, is the same as the multilayer in AF ordering. Following this argument, our samples can be in such an un-coupled random magnetic state at  $H_0$ , as long as the total magnetization is zero.

In the following sections, the two channel model equations (4.9) and (4.10) are used to analyze our data at  $H_0$  and  $H_s$ , respectively. A global fit to selected sets of data is presented in section 5.5.

All Ag layers next to Nb are assumed to have zero resistance under the magnetic fields we use, due to the superconducting proximity effect. The resistance reduction is usually much smaller than our measuring uncertainties. The Nb/Co interface resistance is assumed spin independent. The units we use in our equations are  $f\Omega m^2$  for ARs, n $\Omega$ m for resistivities, and  $\mu$ m for thicknesses, unless noted otherwise.

### 5.4.1 Experimental Results

See Tables 5.17-5.24.

| M    | sample no. | AR    | $R_T(f\Omega m)$ | 1 <sup>2</sup> ) | error <sup>a</sup> % | $\sqrt{[r_0-r_s]r_0}$ |
|------|------------|-------|------------------|------------------|----------------------|-----------------------|
|      |            | $H_0$ | $H_p$            | H,               |                      |                       |
| 1.5  | 337-03     | 8.4   | 8.3              | 8.1              | 5.2 %                | 1.7                   |
| 3.5  | 337-04     | 13.1  | 12.4             | 11.2             | 5.2 %                | 5.0                   |
| 5.5  | 337-05     | 16.6  | 15.3             | 13.1             | 3.1 %                | 7.6                   |
| 10.5 | 350-01     | 29.0  | 25.0             | 20.6             | 3.3 %                | 15.6                  |
| 20.5 | 350-02     | 48.0  | 39.7             | 29.9             | 3.2 %                | 29.5                  |
| 30.5 | 350-03     | 64.6  | 49.8             | 37.3             | 2.7 %                | 42.0                  |
| 60   | 246-01     | 113.9 | 95.4             | 67.2             | 2.9 %                | 73.0                  |
| 60   | 298-01     | 123.9 | 98.8             | 74.6             | 3.2 %                | 78.1                  |
| 60   | 310-01     | 121.8 | 92.7             | 65.6             | 2.3 %                | 82.7                  |
| 60   | 336-01     | 123.2 | 91.4             | 67.6             | 2.8 %                | 82.8                  |
| 60   | 337-01     | 120.3 | 94.7             | 67.1             | 3.3 %                | 81.0                  |
| 60   | 339-01     | 122.0 | 93.7             | 67.4             | 1.8 %                | 81.6                  |

Table 5.17.  $AR_T$  values at three magnetic fields of  $[Co(6nm)/Ag(6nm)] \times M$  samples.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

<sup>a</sup>Errors for M = 60 samples are taken as the area measurement error, like all other samples in other sections. Resistance errors are small for these samples. For samples with smaller resistance, i.e.  $M \leq 30.5$ , both resistance and area errors are taken into account.

Table 5.18.  $AR_T$  values at three magnetic fields of  $[Co(6nm)/Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

| $t_{Ag}(nm)$ | M  | sample no. | A     | $R_T(\mathrm{f}\Omega\mathrm{m})$ | 1 <sup>2</sup> ) | error % | $\sqrt{[r_0-r_s]r_0}$ |
|--------------|----|------------|-------|-----------------------------------|------------------|---------|-----------------------|
|              |    |            | $H_0$ | $H_p$                             | H <sub>s</sub>   |         |                       |
| 2            | 90 | 251-01     | 105.3 | 105.6                             | 104.2            | 5.0 %   | 11.0                  |
| 4            | 72 | 251-03     | 128.8 | 101.4                             | 74.0             | 2.7 %   | 83.9                  |
| 6            | 60 | 246-01     | 113.9 | 95.4                              | 67.2             | 2.9 %   | 73.0                  |
| 6            | 60 | 298-01     | 123.9 | 98.8                              | 74.6             | 3.2 %   | 78.1                  |
| 6            | 60 | 310-01     | 121.8 | 92.7                              | 65.6             | 2.3 %   | 82.7                  |
| 6            | 60 | 336-01     | 123.2 | 91.4                              | 67.6             | 2.8 %   | 82.8                  |
| 6            | 60 | 337-01     | 120.3 | 94.7                              | 67.1             | 3.3 %   | 80.0                  |
| 6            | 60 | 339-01     | 122.0 | 93.7                              | 67.4             | 1.8 %   | 81.6                  |
| 9            | 48 | 251-04     | 93.7  | 80.8                              | 55.6             | 3.4 %   | 59.7                  |
| 12           | 40 | 251-05     | 78.0  | 70.6                              | 50.3             | 1.9 %   | 47.3                  |
| 12           | 40 | 252-01     | 82.0  | 71.7                              | 48.4             | 1.0 %   | 52.5                  |
| 15           | 34 | 256-01     | 79.4  | 69.7                              | 50.4             | 6.9 %   | 48.0                  |
| 18           | 30 | 252-02     | 65.6  | 57.7                              | 41.2             | 4.0 %   | 40.0                  |
| 22           | 26 | 252-03     | 60.1  | 53.4                              | 40.1             | 11.0 %  | 34.7                  |
| 30           | 20 | 336-02     | 45.2  | 40.7                              | 32.0             | 3.5 %   | 24.5                  |
| 35           | 18 | 256-05     | 43.2  | 38.2                              | 31.1             | 2.1 %   | 22.9                  |
| 60           | 11 | 256-02     | 31.8  | 29.4                              | 26.4             | 6.3 %   | 13.1                  |
| 60           | 11 | 338-02     | 30.0  | 27.9                              | 24.7             | 2.5 %   | 12.6                  |

Table 5.19.  $AR_T$  values at three magnetic fields of  $[Co(2nm)/Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

| $t_{Ag}(nm)$ | M  | sample no. | Al    | $R_T(f\Omega m)$ | <sup>2</sup> ) | error % | $\sqrt{[r_0-r_s]r_0}$ |
|--------------|----|------------|-------|------------------|----------------|---------|-----------------------|
|              |    |            | $H_0$ | $H_p$            | H <sub>s</sub> |         | •                     |
| 6            | 90 | 261-01     | 137.1 | 112.7            | 68.6           | 4.1 %   | 96.9                  |
| 6            | 90 | 338-04     | 125.7 | 102.3            | 61.3           | 2.5 %   | 89.9                  |
| 6            | 90 | 340-01     | 132.8 | 108.6            | 65.0           | 4.4 %   | 94.9                  |
| 8            | 72 | 261-02     | 101.2 | 84.4             | 52.3           | 3.0 %   | 70.3                  |
| 12           | 51 | 261-03     | 76.8  | 62.9             | 43.3           | 1.9 %   | 50.7                  |
| 20           | 33 | 261-04     | 56.7  | 47.6             | 36.0           | 4.5 %   | 34.3                  |
| 58           | 12 | 261-05     | 27.8  | 25.7             | 23.2           | 3.7 %   | 11.3                  |

Table 5.20.  $AR_T$  values at three magnetic fields of  $[Co(t)=Ag(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

| t(nm) | M  | sample no. | AR    | $2_T(f\Omega m)$ | n <sup>2</sup> ) | error % | $\sqrt{[r_0-r_s]r_0}$ |
|-------|----|------------|-------|------------------|------------------|---------|-----------------------|
|       |    |            | $H_0$ | $H_p$            | H <sub>s</sub>   |         |                       |
| 6     | 60 | 246-01     | 113.9 | 95.4             | 67.2             | 2.9 %   | 73.0                  |
| 6     | 60 | 298-01     | 123.9 | 98.8             | 74.6             | 3.2 %   | 78.1                  |
| 6     | 60 | 310-01     | 121.8 | 92.7             | 65.6             | 2.3 %   | 82.7                  |
| 6     | 60 | 336-01     | 123.2 | 91.4             | 67.6             | 2.8 %   | 82.8                  |
| 6     | 60 | 337-01     | 120.3 | 94.7             | 67.1             | 3.3 %   | 80.0                  |
| 6     | 60 | 339-01     | 122.0 | 93.7             | 67.4             | 1.8 %   | 81.6                  |
| 8     | 45 | 246-05     | 97.4  | 80.6             | 58.9             | 3.9 %   | 61.2                  |
| 12    | 30 | 337-02     | 74.5  | 61.6             | 48.2             | 2.8 %   | 44.3                  |
| 15    | 24 | 246-09     | 69.6  | 61.0             | 49.9             | 3.7 %   | 37.0                  |
| 18    | 20 | 336-03     | 62.1  | 53.2             | 45.2             | 2.6 %   | 32.4                  |
| 20    | 18 | 260-01     | 62.5  | 53.1             | 47.0             | 4.1 %   | 31.2                  |
| 30    | 12 | 260-02     | 49.6  | 43.6             | 40.8             | 4.5 %   | 20.9                  |
| 50    | 7  | 260-03     | 39.7  | 38.2             | 35.8             | 4.2 %   | 12.6                  |

| $t_{Co}(nm)$ | M  | sample no. | $AR_T(\mathrm{f}\Omega\mathrm{m}^2)$ |       |      | error % | $\sqrt{[r_0-r_s]r_0}$ |
|--------------|----|------------|--------------------------------------|-------|------|---------|-----------------------|
|              |    |            | $H_0$                                | $H_p$ | H,   |         | ·                     |
| 1.5          | 96 | 276-01     | 136.7                                | 113.4 | 68.4 | 3.7 %   | 96.6                  |
| 2            | 90 | 261-01     | 137.1                                | 112.7 | 68.6 | 4.1 %   | 96.9                  |
| 2            | 90 | 338-04     | 125.7                                | 102.3 | 61.3 | 2.5 %   | 89.9                  |
| 2            | 90 | 340-01     | 132.8                                | 108.6 | 65.0 | 4.4 %   | 94.9                  |
| 4            | 72 | 276-02     | 121.8                                | 99.0  | 61.5 | 3.0 %   | 85.7                  |
| 6            | 60 | 246-01     | 113.9                                | 95.4  | 67.2 | 2.9 %   | 73.0                  |
| 6            | 60 | 298-01     | 123.9                                | 98.8  | 74.6 | 3.2 %   | 78.1                  |
| 6            | 60 | 310-01     | 121.8                                | 92.7  | 65.6 | 2.3 %   | 82.7                  |
| 6            | 60 | 336-01     | 123.2                                | 91.4  | 67.6 | 2.8 %   | 82.8                  |
| 6            | 60 | 337-01     | 120.3                                | 94.7  | 67.1 | 3.3 %   | 80.0                  |
| 6            | 60 | 339-01     | 122.0                                | 93.7  | 67.4 | 1.8 %   | 81.6                  |
| 9            | 48 | 339-02     | 102.6                                | 73.9  | 61.6 | 1.6 %   | 64.8                  |
| 12           | 40 | 276-03     | 89.7                                 | 64.9  | 58.4 | 3.0 %   | 53.0                  |
| 12           | 40 | 338-03     | 82.1                                 | 67.8  | 61.4 | 2.6 %   | 41.2                  |
| 30           | 20 | 340-02     | 59.5                                 | 59.4  | 58.2 | 3.3 %   | 8.8                   |
| 60           | 11 | 276-04     | 52.2                                 | 52.9  | 50.1 | 4.2 %   | 10.3                  |

Table 5.21.  $AR_T$  values at three magnetic fields of  $[Co(t)/Ag(6nm)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

Table 5.22.  $AR_T$  values at three magnetic fields of  $[Co(6nm)/Ag(t)] \times 60$  samples.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

| $t_{Ag}(nm)$ | sample no. | Al    | $R_T(f\Omega m)$ | <sup>2</sup> ) | error % | $\sqrt{[r_0-r_s]r_0}$ |
|--------------|------------|-------|------------------|----------------|---------|-----------------------|
|              |            | $H_0$ | $H_p$            | H <sub>s</sub> |         | •                     |
| 4            | 300-01     | 105.8 | 92.0             | 79.0           | 2.3 %   | 53.2                  |
| 4            | 338-05     | 106.2 | 81.7             | 60.5           | 2.1 %   | 69.6                  |
| 4            | 350-04     | 92.5  | 76.7             | 62.0           | 2.3 %   | 53.2                  |
| 5            | 310-03     | 115.7 | 89.1             | 63.3           | 3.1 %   | 77.8                  |
| 6            | 246-01     | 113.9 | 95.4             | 67.2           | 2.9 %   | 73.0                  |
| 6            | 298-01     | 123.9 | 98.8             | 74.6           | 3.2 %   | 78.1                  |
| 6            | 310-01     | 121.8 | 92.7             | 65.6           | 2.3 %   | 82.7                  |
| 6            | 336-01     | 123.2 | 91.4             | 67.6           | 2.8 %   | 82.8                  |
| 6            | 337-01     | 120.3 | 94.7             | 67.1           | 3.3 %   | 80.0                  |
| 6            | 339-01     | 122.0 | 93.7             | 67.4           | 1.8 %   | 81.6                  |
| 9            | 310-04     | 122.3 | 102.8            | 69.1           | 1.9 %   | 80.6                  |
| 12           | 300-03     | 121.4 | 104.2            | 70.8           | 2.6 %   | 78.4                  |
| 12           | 338-07     | 110.1 | 94.1             | 64.5           | 2.8 %   | 70.9                  |
| 15           | 308-02     | 116.9 | 101.4            | 66.3           | 1.7 %   | 76.9                  |
| 15           | 350-07     | 111.9 | 96.8             | 63.8           | 4.3 %   | 73.4                  |
| 20           | 338-06     | 113.2 | 95.0             | 69.4           | 2.1 %   | 70.4                  |
| 30           | 350-08     | 108.4 | 94.0             | 66.8           | 4.2 %   | 67.1                  |

| $t_{Co}(\text{nm})$ | sample no. | A     | $R_T(f\Omega m)$ | 1 <sup>2</sup> ) | error % | $\sqrt{[r_0-r_s]r_0}$ |
|---------------------|------------|-------|------------------|------------------|---------|-----------------------|
|                     |            | $H_0$ | $H_p$            | H,               |         |                       |
| 2                   | 297-01     | 95.6  | 79.9             | 50.9             | 1.9 %   | 65.4                  |
| 6                   | 246-01     | 113.9 | 95.4             | 67.2             | 2.9 %   | 73.0                  |
| 6                   | 298-01     | 123.9 | 98.8             | 74.6             | 3.2 %   | 78.1                  |
| 6                   | 310-01     | 121.8 | 92.7             | 65.6             | 2.3 %   | 82.7                  |
| 6                   | 336-01     | 123.2 | 91.4             | 67.6             | 2.8 %   | 82.8                  |
| 6                   | 337-01     | 120.3 | 94.7             | 67.1             | 3.3 %   | 80.0                  |
| 6                   | 339-01     | 122.0 | 93.7             | 67.4             | 1.8 %   | 81.6                  |
| 9                   | 297-03     | 135.0 | 104.4            | 83.2             | 2.2 %   | 83.7                  |
| 12                  | 352-01     | 131.1 | 106.0            | 96.1             | 2.1 %   | 67.8                  |
| 15                  | 308-01     | 141.6 | 124.4            | 113.0            | 2.7 %   | 63.6                  |
| 18                  | 352-02     | 127.3 | 124.8            | 120.0            | 5.0 %   | 30.4                  |
| 20                  | 298-03     | 132.7 | 132.0            | 127.8            | 2.5 %   | 23.4                  |

Table 5.23.  $AR_T$  values at three magnetic fields of  $[Co(t)/Ag(6nm)] \times 60$  samples.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

Table 5.24.  $AR_T$  values at three magnetic fields of  $[Co(6nm)/AgSn(t)] \times M$  samples. Total sample thicknesses are as close as possible to 720nm.  $r_0$  and  $r_s$  are  $AR_T$  at  $H_0$  and  $H_s$ , respectively.

| $t_{AgSn}(nm)$ | M  | sample no. | $AR_T(\mathrm{f}\Omega\mathrm{m}^2)$ |       |                | error % | $\sqrt{[r_0-r_s]r_0}$ |
|----------------|----|------------|--------------------------------------|-------|----------------|---------|-----------------------|
|                |    |            | $H_0$                                | $H_p$ | H <sub>s</sub> |         |                       |
| 2              | 90 | 262-01     | 141.8                                | 142.0 | 140.7          | 2.1 %   | 12.5                  |
| 4              | 72 | 262-02     | 179.5                                | 156.8 | 138.6          | 4.6 %   | 20.5                  |
| 6              | 60 | 262-03     | 182.9                                | 165.7 | 145.8          | 3.6 %   | 44.8                  |
| 6              | 60 | 277-04     | 169.7                                | 151.3 | 132.1          | 2.5 %   | 53.7                  |
| 12             | 40 | 277-02     | 165.8                                | 158.6 | 148.4          | 2.6 %   | 79.8                  |
| 15             | 34 | 262-04     | 167.9                                | 163.3 | 155.9          | 2.4 %   | 82.4                  |
| 35             | 18 | 262-05     | 146.1                                | 144.7 | 143.2          | 3.3 %   | 85.7                  |
| 60             | 11 | 277-03     | 128.1                                | 127.5 | 127.1          | 4.4 %   | 12.5                  |



Figure 5.9. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(6nm)/Ag(6nm)] \times M$  samples. Best fit lines:  $y = (7.0 \pm 0.3) + (1.95 \pm 0.02)x$  for  $H_0$ ,  $y = (7.8 \pm 0.3) + (1.47 \pm 0.02)x$  for  $H_p$ . If we do a linear fit for  $H_s$ ,  $y = (8.0 \pm 0.3) + (1.02 \pm 0.01)x$ .  $\chi^2 = 18.5$ , 26.1, 33.0, respectively.



Figure 5.10.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M - 1 for  $[Co(6nm)/Ag(6nm)] \times M$  samples. Best fit line:  $y = (0.9 \pm 0.2) + (1.36 \pm 0.01)x$ .  $\chi^2 = 29.0$ .



Figure 5.11. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(6nm)/Ag(t)] \times M$ samples. All samples have total thickness 720nm. Best fit lines:  $y = (11.8 \pm 0.8) + (1.81 \pm 0.02)x$  for  $H_0$ ,  $y = (15.0 \pm 0.7) + (1.39 \pm 0.02)x$  for  $H_p$ . If we do a linear fit for  $H_s$ ,  $y = (16.1 \pm 0.6) + (0.85 \pm 0.02)x$ .  $\chi^2 = 23.8$ , 34.2, 22.5, respectively.



Figure 5.12.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1 for  $[Co(6nm)/Ag(t)] \times M$  samples. All samples have total thickness 720nm. Best fit line:  $y = (-0.75 \pm 0.35) + (1.35 \pm 0.01)x$ .  $\chi^2 = 53.2$ .



Figure 5.13. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(2nm)/Ag(t)] \times M$ samples. All samples have total thickness 720nm. Best fit lines:  $y = (13.8 \pm 1.2) + (1.27 \pm 0.03)x$  for  $H_0$ ,  $y = (14.5 \pm 1.1) + (1.00 \pm 0.02)x$  for  $H_p$ . If we do a linear fit for  $H_s$ ,  $y = (17.9 \pm 0.9) + (0.51 \pm 0.02)x$ .  $\chi^2 = 6.3$ , 7.5, 8.2, respectively.



Figure 5.14.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1 for  $[Co(2nm)/Ag(t)] \times M$  samples. All samples have total thickness 720nm. Best fit line:  $y = (0.10 \pm 0.52) + (1.02 \pm 0.02)x$ .  $\chi^2 = 5.7$ .



Figure 5.15. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(t)=Ag(t)] \times M$ samples. All samples have total thickness 720nm. Best fit lines:  $y = (34.9 \pm 1.8) + (1.45 \pm 0.04)x$  for  $H_0$ ,  $y = (34.7 \pm 1.5) + (1.00 \pm 0.03)x$  for  $H_p$ . If we do a linear fit for  $H_s$ ,  $y = (35.0 \pm 1.2) + (0.55 \pm 0.03)x$ .  $\chi^2 = 9.1$ , 8.6, 18.1, respectively.



Figure 5.16.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M - 1 for  $[Co(t)=Ag(t)] \times M$  samples. All samples have total thickness 720nm. Best fit line:  $y = (10.1 \pm 1.0) + (1.19 \pm 0.03)x$ .  $\chi^2 = 17.2$ .



Figure 5.17. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(t)/Ag(6nm)] \times M$  samples. All samples have total thickness 720nm.



Figure 5.18.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1 for  $[Co(t)/Ag(6nm)] \times M$  samples. All samples have total thickness 720nm.



Figure 5.19. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(6nm)/Ag(t)] \times 60$  samples.



Figure 5.20. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(t)/Ag(6nm)] \times 60$  samples.



Figure 5.21.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus Co thickness for  $[Co(t)/Ag(6nm)] \times 60$  samples.



Figure 5.22. Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$ , for  $[Co(6nm)/AgSn(t)] \times M$  samples. All samples have total thickness 720nm.



Figure 5.23. Differences of area times CPP resistances at  $H_0$  and  $H_s$  for  $[Co(6nm)/AgSn(t)] \times M$  samples, circles, and of  $[Co(6nm)/Ag(t)] \times M$  samples, diamonds. All samples have total thickness 720nm.



Figure 5.24.  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M - 1 for  $[Co(6nm)/AgSn(t)] \times M$  samples. All samples have total thickness 720nm. The solid line is the best fit to Co/AgSn data  $y = (-3.45 \pm 0.62) + (1.44 \pm 0.03)x$ .  $\chi^2 = 2.3$ . The dotted line is the best fit for equivalent Co/Ag samples (Figure 5.12).

#### 5.4.2 Co(6nm)/Ag(6nm) with increasing bilayer number M

Samples with Co and Ag thicknesses both fixed at 6nm, but various bilayer numbers, and thus various total thicknesses, are listed in Table 5.17, where half a bilayer means an extra layer of Co, e.g. M = 1.5 means the sample consists of two layers of Co with one layer of Ag in between.

Since, in this set of samples, we are adding in more and more Co/Ag bilayers of the same thickness, it is like putting more of exactly the same resistor in series, assuming that the crystalline structure does not change with sample thickness. However, preliminary results of Transmission Electron Microscope images of our samples show that the layering is uniform for the first few bilayers, but when the samples get thicker, columnar growth begins, leading to a more complex structure. The layering between different columns may not always register well, but within each column, the layering is still uniform. We thus have to test whether the series resistance model is compatible with our samples. Furthermore, the elastic mean free paths estimated from a free electron model are around 40nm for our sputtered Co films and around 100nm for Ag films (see section 3.2). The total sample thickness, which ranges from 18nm to 720nm, goes from smaller than the shorter mean free path, to much larger than the longer one. It is important to see if there is any evidence of mean free path effects on our data.

From equations (4.9)(4.10), the two channel series resistance model gives, with M denoting the Co layer number for this set:

$$\begin{aligned} AR_T(H_0) &= 2AR_{Nb/Co} + \rho_{Co}^* t_{Co} + (M-1) \left[ \rho_{Co}^* t_{Co} + \rho_{Ag} t_{Ag} + 2AR_{Co/Ag}^* \right] \\ \sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)} &= \beta \rho_{Co}^* t_{Co} + (M-1) \left[ \beta \rho_{Co}^* t_{Co} + \gamma 2AR_{Co/Ag}^* \right]. \end{aligned}$$

Since  $\rho_{Co}^* t_{Co}$  is only of order 0.6 f $\Omega$ m<sup>2</sup>, these equations predict that: (1) a plot of AR at  $H_0$  versus M - 1 should give a straight line, and the intercept with the y-axis

should be close to the independently measured  $2AR_{Nb/Co} = 6.1 \,\mathrm{f\Omega m^2}$ ; and (2) a plot of  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M - 1 should be a straight line, going closely through the origin.

Figures 5.9 and 5.10 show our data. The solid lines are the weighted least square fits. The arrow pointing to the y-axes is the intercepts predicted by the two channel model. We see that the two channel model fits our data very well. A weighted least square fit to the data yields:

$$\begin{aligned} 2AR_{Nb/Co} &+ 0.006\rho_{Co}^* = 7.00 \pm 0.29 \\ 0.006(\rho_{Co}^* + \rho_{Ag}) + 2AR_{Co/Ag}^* = 1.95 \pm 0.02 \\ \beta\rho_{Co}^* &= 157 \pm 34 \\ 0.006\beta\rho_{Co}^* + \gamma 2AR_{Co/Ag}^* = 1.36 \pm 0.01 \quad , \end{aligned}$$

 $\chi^2 = 18.5, 29.0$ , respectively. These equations, together with equations we shall find in the following sections, can be used to solve for  $2AR_{Nb/Co}$ ,  $\rho_{Ag}$ ,  $\rho_{Co}^*$ ,  $AR_{Co/Ag}^*$ ,  $\beta$ , and  $\gamma$ . We shall see, however, that not all the unknowns can be solved consistently. Instead of solving these equations, we will do a global fit to our data to get these quantities directly.  $2AR_{Nb/Co}$  and  $\rho_{Ag}$  can then be compared with independent measurement.

The fact that the data at  $H_0$  and  $H_p$  fall closely on straight lines means that the model works for a wide range of total sample thicknesses; the quality of our samples does not affect the CPP resistances for total sample thickness from ten nanometers to almost a micron. There is no obvious effect we can associate with the mean free paths.

## 5.4.3 Fixed Co(6nm) various Ag thicknesses with total sample thickness fixed at 720nm

Samples were made with the Co thickness fixed at 6nm and various Ag thicknesses, keeping the total thickness as close as possible to 720nm, consistent with integer bilayer numbers. Since the Co thickness is fixed, we expect the magnetic structure to be the same. The results are listed in Table 5.18.

With  $t_{Ag} = \frac{t_T}{M} - t_{Co}$ , the two channel model gives:

$$\begin{aligned} AR_T(H_0) &= 2AR_{Nb/Co} + \rho_{Co}^* t_{Co} + (1 - \frac{1}{M})\rho_{Ag} t_T \\ &+ (M - 1) \left[ (\rho_{Co}^* - \rho_{Ag}) t_{Co} + 2AR_{Co/Ag}^* \right] \\ \sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)} &= \beta \rho_{Co}^* t_{Co} + (M - 1) \left[ \beta \rho_{Co}^* t_{Co} + \gamma 2AR_{Co/Ag}^* \right]. \end{aligned}$$

Both quantities are linear functions of M - 1 if we ignore the term  $-\frac{1}{M}\rho_{Ag}t_T$ . This term has a value of  $0.7 \text{ f}\Omega\text{m}^2$  for the smallest M = 11, which is less than the error of the data. If we ignore this term and extrapolate our  $H_0$  data back to M - 1 = 0, the intercept should be close to the independently measured value,  $2AR_{Nb/Co} + \rho_{Ag}t_T =$  $13.3 \text{ f}\Omega\text{m}^2$ . Furthermore, the straight line in the  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$ versus M - 1 graph should go almost through the origin.

In section 5.1.3 we argued that when Ag is 4nm or less, there is ferromagnetic ordering between Co in our samples. Figures 5.11, 5.12 show our data and the least square fit lines to  $t_{Ag} \ge 6$ nm data. The arrow shows the predicted intercept from the two channel model. A weighted least square fit to  $t_{Ag} \ge 6$ nm data gives:

$$2AR_{Nb/Co} + 0.006\rho_{Co}^* + 0.72\rho_{Ag} = 11.8 \pm 0.8$$
$$0.006(\rho_{Co}^* - \rho_{Ag}) + 2AR_{Co/Ag}^* = 1.81 \pm 0.02$$
$$\beta \rho_{Co}^* = -124 \pm 59$$

$$0.006\beta \rho_{Co}^* + \gamma 2AR_{Co/Ag}^* = 1.35 \pm 0.02$$

 $\chi^2 = 23.8$ , 53.2, respectively. The data fall closely to straight lines in both figures, as required by the model. And all intercepts are close to the predicted values. There is an unphysical negative intercept in Figure 5.12. Considering the reproducibility of our data, as represented by the samples with 6nm Ag, this slight disagreement can be attributed to data fluctuation.

## 5.4.4 Fixed Co(2nm) various Ag thicknesses with total sample thickness fixed at 720nm

To see if a different Co thickness would show different behavior, another set of samples were made with the Co thickness fixed at 2nm and various Ag thicknesses, keeping the total thickness as close as possible to 720nm, consistent with integer bilayer numbers. The results are listed in Table 5.19.

The two channel model equations for this set of samples are exactly the same as those in the previous section. Notice that two channel model predicts very close intercepts of  $AR_T(H_0)$  versus M-1 between these two sets of data, and the straight line in  $\sqrt{[AR_T(H_0) - AR_T(H_*)]AR_T(H_0)}$  versus M-1 graph should again go almost through the origin.

Figures 5.13, 5.14 show the data with weighted least square fit lines and predicted intercept position. A weighted least square fit gives:

$$2AR_{Nb/Co} + 0.002\rho_{Co}^* + 0.72\rho_{Ag} = 13.8 \pm 1.2$$
  
$$0.002(\rho_{Co}^* - \rho_{Ag}) + 2AR_{Co/Ag}^* = 1.27 \pm 0.03$$
  
$$\beta \rho_{Co}^* = 48 \pm 260$$
  
$$0.002\beta \rho_{Co}^* + \gamma 2AR_{Co/Ag}^* = 1.02 \pm 0.02 \quad ,$$

,

 $\chi^2 = 6.3$ , 5.7, respectively. The data for this set of samples also fall along straight lines and the best fit lines have intercepts that match the predictions very well.

# 5.4.5 Equal but varying Co and Ag thicknesses with total sample thickness fixed at 720nm

For this set of samples, the Co and Ag layer thicknesses are kept the same and varied together, and the total sample thicknesses are kept as close as possible to 720nm, consistent with integer bilayer numbers. The results are listed in Table 5.20.

The last two samples listed in the table,  $t_{Co} = 30,50$  nm, display more complex structures in their resistance and magnetic moment versus field curves (see Figure 3.3). It is clear that the magnetic properties are different in these thick Co layers. We thus exclude them from our numerical analysis.

With  $t_{Co} = t_{Ag} = \frac{t_T}{2M}$ , the two channel model gives:

$$AR_{T}(H_{0}) = 2AR_{Nb/Co} + \frac{1}{2}(\rho_{Co}^{*} + \rho_{Ag})t_{T} - \frac{1}{2M}\rho_{Ag}t_{T} + (M-1)2AR_{Co/Ag}^{*}$$

$$\sqrt{[AR_{T}(H_{0}) - AR_{T}(H_{s})]AR_{T}(H_{0})} = \beta\rho_{Co}^{*}\frac{t_{T}}{2} + (M-1)\gamma 2AR_{Co/Ag}^{*}$$

Again, we ignore the  $-\frac{1}{2M}\rho_{Ag}t_T$  term because it is small  $(0.2 \text{ f}\Omega \text{m}^2 \text{ for the smallest} M)$ . Both quantities are linear functions of M - 1. Most importantly, we can get directly from the slope the quantities  $2AR^*_{Co/Ag}$  and  $\gamma$  which can not be independently measured.  $\gamma$  can be compared with the  $\beta$  obtained somewhere else to determine the relative importance of interface and bulk contributions for Giant MR.

Figures 5.15, 5.16 show our data and weighted least square fit lines. A weighted least square fit has results:

$$2AR_{Nb/Co} + 0.36(\rho_{Co}^* + \rho_{Ag}) = 34.9 \pm 1.8$$
$$AR_{Co/Ag}^* = 0.73 \pm 0.04$$

$$eta 
ho_{Co}^* = 28.0 \pm 2.7$$
  
 $\gamma A R_{Co/Ag}^* = 0.59 \pm 0.01$  ,

 $\chi^2 = 9.1, 17.2$ , respectively. Using these equations, together with those we obtained in the previous three sections, we have more than enough equations to determine the unknowns. By doing so we overweigh those samples repeatedly listed in different sets. Instead, we will do a global fit to these four sets of data in section 5.5, without double counting any samples.

## 5.4.6 Fixed Ag(6nm) various Co thicknesses with total sample thickness fixed at 720nm

Samples were made with Ag thickness fixed at 6nm, various Co thicknesses but keeping the total thicknesses as close as possible to 720nm, consistent with integer bilayer numbers. In this set of samples, we expect to see effects due primarily to changing Co thickness. Results are listed in Table 5.21.

With  $t_{Co} = \frac{t_T}{M} - t_{Ag}$ , the two channel model gives:

$$\begin{aligned} AR_T(H_0) &= 2AR_{Nb/Co} + \rho_{Co}^* t_T + \rho_{Co}^* t_{Ag} \\ &+ (M - 1) \left[ (-\rho_{Co}^* + \rho_{Ag}) t_{Ag} + 2AR_{Co/Ag}^* \right] \\ \sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)} \\ &= \beta \rho_{Co}^* (t_T - t_{Ag}) + (M - 1) \left[ -\beta \rho_{Co}^* t_{Ag} + \gamma 2AR_{Co/Ag}^* \right] \end{aligned}$$

Both quantities are linear in M-1 with finite intercepts.

We can see from Figures 5.17, 5.18 that the data are more scattered than previous sets. The data at  $H_p$  for 9nm Co and thicker samples, i.e.  $M \leq 48$ , are much closer to the saturation data than the data for thinner samples. Even for data at  $H_0$ , there is a trend for the data of thicker Co samples to fall below the straight line extrapolated from thinner Co samples. We explain this as due to the properties of the Co layers changing as the Co gets thicker (M gets smaller). Instead of fitting the data, we shall use the global fit values from other sets of data to compare these data with predictions from the two channel model.

## 5.4.7 Fixed Co(6nm) various Ag thicknesses with bilayer number M = 60

Samples with Co thickness fixed at 6nm bilayer number fixed at 60, and various Ag thicknesses, thus various total thicknesses, were made with the expectation that the ARs at  $H_0$  would be proportional to the Ag thickness. The results are listed in Table 5.22.

The ARs of three samples with 4nm Ag are scattered. We believe this is because the Ag layer is so thin that adjacent Co layers show some magnetic ordering, and the detailed ordering is sensitive to thickness variations (see section 5.1.3).

The two channel model gives, in this case:

$$AR_{T}(H_{0}) = 2AR_{Nb/Co} + M\rho_{Co}^{*}t_{Co} + (M-1)\rho_{Ag}t_{Ag} + 2(M-1)AR_{Co/Ag}^{*}$$

$$\sqrt{[AR_{T}(H_{0}) - AR_{T}(H_{s})]AR_{T}(H_{0})} = M\beta\rho_{Co}^{*}t_{Co} + 2(M-1)\gamma AR_{Co/Ag}^{*}$$

A plot of AR at  $H_0$  versus  $t_{Ag}$  should be a straight line with slope  $59\rho_{Ag}$  and  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  for all samples should have the same value.

Figure 5.19 shows our data. A weighted least square fit to  $H_0$  data with  $t_{Ag} \ge 5$ nm shows

$$\begin{aligned} 2AR_{Nb/Co} &+ 0.36\rho_{Co}^* + 118AR_{Co/Ag}^* = 123.8 \pm 1.7 \\ \rho_{Ag} &= -8.8 \pm 2.3 \quad , \end{aligned}$$

 $\chi^2 = 18.5$ . We find a slightly negative slope, instead of the slightly positive one expected. However, considering that the variation of nominally identical samples is about 5%, this set of data is not incompatible with the model. Since the mean free path estimated from the free electron model for our sputter deposited Ag film is about 80-100nm, this set of data simply proves that varying the Ag layer thicknesses from 5nm to 30nm does not show any marked dependence on the mean free path.

The average and standard deviation (n-1) of  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$ for  $t_{Ag} \ge 5$ nm samples gives:

$$0.36\beta \rho_{Co}^* + 118\gamma A R_{Co/Ag}^* = 76.7 \pm 5.0$$

### 5.4.8 Various Co thicknesses with fixed Ag(6nm), bilayer number M = 60

Samples with various Co thicknesses, fixed Ag thickness at 6nm, bilayer number fixed at 60, and thus various total thicknesses, were made to see if the ARs at  $H_0$  would be proportional to the Co thickness. The results are listed in Table 5.23.

The ARs at  $H_0$  and  $H_p$  of samples with  $t_{Co} \leq 9$ nm show linear behavior with Co thicknesses, but for  $t_{Co} \geq 12$ nm they fall closer to the saturation values and for  $t_{Co} = 20$ nm they fall almost on the saturation value.

A possible reason for this behavior is that as the Co layer gets thicker, some magnetic structure develops within the layer, thus smearing out the giant MR effect. More study on thicker Co samples is needed to fully understand their behavior.

The two channel model gives:

$$AR_{T}(H_{0}) = 2AR_{Nb/Co} + M\rho_{Co}^{*}t_{Co} + (M-1)\rho_{Ag}t_{Ag} + 2(M-1)AR_{Co/Ag}^{*}$$

$$\sqrt{[AR_{T}(H_{0}) - AR_{T}(H_{s})]AR_{T}(H_{0})} = M\beta\rho_{Co}^{*}t_{Co} + 2(M-1)\gamma AR_{Co/Ag}^{*}$$

Figures 5.20, 5.21 show our data. A weighted least square fit to  $t_{Co} \leq 9$ nm samples shows:

$$\begin{split} 2AR_{Nb/Co} &+ & 0.354\rho_{Ag} + 118AR^*_{Co/Ag} = 84.6 \pm 2.5\\ \rho^*_{Co} &= & 98.8 \pm 7.3\\ \gamma AR^*_{Co/Ag} &= & 0.515 \pm 0.014\\ \beta \rho^*_{Co} &= & 49.8 \pm 4.9 \quad , \end{split}$$

 $\chi^2 = 7.5, 25.1$ , respectively.

## 5.4.9 Fixed Co(6nm) various AgSn thicknesses with total sample thickness fixed at 720nm

To test the mean free path effect in the normal metal layers, 4at.% of Sn was added to the Ag to make AgSn alloy spacer layers. Multilayers were made with Co thickness fixed at 6nm and various AgSn thicknesses, with the total thickness as close as possible to 720nm, consistent with integer bilayer numbers. The results are listed in Table 5.24.

Figure 5.22 shows the data. If we plot  $AR_T(H_0) - AR_T(H_s)$  versus M - 1, we see that samples with  $M \leq 60$  ( $t_{AgSn} \geq 6$ nm) fall on a smooth curve and the two samples with M = 72,90 ( $t_{AgSn} = 2$ nm,4nm) have smaller values than the extended curve. As for Co/Ag, we assume that the samples with  $t_{AgSn} = 2$ nm,4nm are (partially) ferromagnetically ordered. Thus we can say that the presence of Sn does not change the magnetic ordering significantly.

The two channel model gives:

$$AR_{T}(H_{0}) = 2AR_{Nb/Co} + \rho_{Co}^{*}t_{Co} + (1 - \frac{1}{M})\rho_{AgSn}t_{T}$$

$$+ (M - 1) \left[ (\rho_{Co}^{*} - \rho_{AgSn}) t_{Co} + 2AR_{Co/Ag}^{*} \right]$$

$$AR_{T}(H_{0}) - AR_{T}(H_{s}) = \frac{\left[ M\beta \rho_{Co}^{*} t_{Co} + (M - 1)\gamma 2AR_{Co/Ag}^{*} \right]^{2}}{AR_{T}(H_{0})}$$

$$\sqrt{[AR_{T}(H_{0}) - AR_{T}(H_{s})]AR_{T}(H_{0})} = \beta \rho_{Co}^{*} t_{Co} + (M - 1) \left[ \beta \rho_{Co}^{*} t_{Co} + \gamma 2AR_{Co/Ag}^{*} \right]$$

Taking a closer look at the equation for  $AR_T(H_0) - AR_T(H_o)$ , we note that the denominator,  $AR_T(H_0)$ , is linear in M and the numerator is quadratic in M. When M is large, the constant term in the denominator is negligible, and  $AR_T(H_0) - AR_T(H_o)$  has a linear dependence on M. When M is small, the constant term dominates the denominator and  $AR_T(H_0) - AR_T(H_o)$  has a quadratic dependence on M. The position of the turning point which separates these two behavior depends on the constant term,  $2AR_{Nb/Co} + \rho_{Co}^* t_{Co} + \rho_{AgSn} t_T$ . Figure 5.23 shows  $AR_T(H_0) - AR_T(H_o)$  versus M - 1 for both AgSn and equivalent pure Ag (section 5.4.3) data. For pure Ag,  $\rho_{Ag}$  is small, and the turning point is at very small M. Our data show almost a straight line. On the other hand, AgSn has a resistivity about 20 times larger than pure Ag, causing the turning point to shift to much larger M. We can see this effect clearly from the plot.

The quantity  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  does not depend on the resistivity of the spacer layer. Thus the two channel model predicts the same dependence on M - 1 for pure Ag and AgSn samples. Figure 5.24 shows our Co/AgSn data. A weighted least square fit to AgSn data along gives:

$$\beta \rho_{Co}^* = -576 \pm 102$$
$$0.006 \beta \rho_{Co}^* + \gamma 2 A R_{Co/Ag}^* = 1.44 \pm 0.03$$

,

 $\chi^2 = 2.3$ . These values are comparable to those in section 5.4.3. The solid line in Figure 5.24 is the best fit to Co/AgSn data, and the dotted line is the best fit for the equivalent Co/Ag samples (Figure 5.12).

We conclude that the presence of 4at.% Sn in the Ag layers does not change the coupling between Co layers. Although the mean free paths differ by a factor of about 20, the two channel model works well in interpreting both sets of data.

#### 5.5 Global Fit

In the previous sections we presented the CPP ARs of seven sets of Co/Ag samples and one set of Co/AgSn samples. In the case of pure Ag, we have  $AR_{Nb/Co}$ ,  $\rho_{Ag}$ ,  $\rho_{Co}^*$ ,  $AR_{Co/Ag}^*$ ,  $\beta$ , and  $\gamma$ , six unknowns in the two channel model. When we fit those sets of data independently, we have more than enough equations to determine six unknowns. However, we sometimes got unphysical numbers, like negative resistivity, and the best fit numbers in different sets for the same quantity did not always agree with each other to within their mutual uncertainties. Instead of solving the best fit equations from different sections to get the unknowns, we present a global fit in this section.

Not all of our sample sets can be completely described by the two channel model. To do a reasonable global fit, we have to limit ourselves to certain sets of samples. From: (a) section 5.4.3, we know that there is (partial) magnetic ordering in our samples when Ag is 4nm or less; (b) sections 5.4.6 and 5.4.8, we know that when the Ag thickness is fixed at 6nm, variable Co layers probably have different magnetic properties when the Co thickness is larger than 12nm; (c) section 5.4.5, we know that when Co and Ag layers are kept equal and varied, samples with thicknesses 30 and 50nm show structure in their resistance and magnetization curves (see Figure 3.3). We thus limit ourselves with the criteria:

- t<sub>Ag</sub> ≥6nm. Experimental results show no obvious ferromagnetic ordering between Co layers.
- $t_{Co} \leq t_{Ag}$  but  $t_{Co} \leq 20$  nm. Where no complex behavior of resistance or magnetization has been seen.

In references [163, 164], we chose three sets of Co/Ag samples for our global fit:

1. Section 5.4.3. Fixed Co thickness at 6nm, total thickness at 720nm. Exclude

samples with  $t_{Ag} < 6nm$ , where ferromagnetic ordering is present.

- 2. Section 5.4.4. Fixed Co thickness at 2nm, total thickness at 720nm.
- 3. Section 5.4.5. Co and Ag thicknesses kept equal but varying, fixed total thickness at 720nm. Exclude two samples  $t_{Co} = 30, 50$ nm.

Since these three sets of samples are not enough to determine all the unknowns in the two channel model, we used the independently measured value for  $AR_{Nb/Co}$ . The results are listed later in this section, where they can be compared with those found from the global fit we present here.

For our global fit, we add in one more set of samples:

4. Section 5.4.2. Fixed Co, Ag thicknesses at 6nm, various M.

With these four sample sets, we can determine all the unknowns from a global fit. The  $[Co(6nm)/Ag(6nm)] \times 60$  samples can be put in any category where they appear but should not be repeated. Listing these samples repeatedly, as we did in previous sections, will cause them to be weighed more than the others. All of the uncertainties in this section have been adjusted so that the reduced  $\chi^2$  is equal to the degrees of freedom as described below. The degrees of freedom equal the number of data points minus the number of unknowns in the fit. We have 34 data points, 4 unknowns in the  $H_0$  fit, and 2 unknowns in the  $H_s$  fit. Thus the degrees of freedom are 30 for the  $H_0$ fit and 32 for the  $H_s$  fit. Uncertainties have been multiplied by  $\sqrt{\chi^2/30}$  and  $\sqrt{\chi^2/32}$ in the  $H_0$  and the  $H_s$  fits, respectively.

From the two channel model, we have

$$AR_{T}(H_{0}) = 2AR_{Nb/Co} + M\rho_{Co}^{*}t_{Co} + (M-1)\rho_{Ag}t_{Ag} + 2(M-1)AR_{Co/Ag}^{*}(5.2) + (M-1)\rho_{Ag}t_{Ag} + 2(M-1)\gamma 2AR_{Co/Ag}^{*}(5.2) - \frac{\left[M\beta\rho_{Co}^{*}t_{Co} + (M-1)\gamma 2AR_{Co/Ag}^{*}\right]^{2}}{AR_{T}(H_{0})} (5.3)$$

$$\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)} = M\beta \rho_{C_o}^* t_{C_o} + (M-1)\gamma 2AR_{C_o/Ag}^*.$$
(5.4)

At  $H_0$ , there are four unknowns:  $AR_{Nb/Co}$ ,  $\rho_{Co}^*$ ,  $\rho_{Ag}$ , and  $AR_{Co/Ag}^*$ . A global fit, with the same principle as a weighted least square fit (see appendix D for details), gives:

$$2AR_{Nb/Co} = 6.9 \pm 0.6 \, \text{f}\Omega \text{m}^2$$
  

$$\rho_{Co}^* = 100 \pm 6 \, \text{n}\Omega \text{m}$$
  

$$\rho_{Ag} = 7.3 \pm 1.9 \, \text{n}\Omega \text{m}$$
  

$$AR_{Co/Ag}^* = 0.60 \pm 0.02 \, \text{f}\Omega \text{m}^2$$

 $\chi^2$  was 111.6 for degrees of freedom 30, before the uncertainties were adjusted.

There are two ways to fit the  $H_s$  data: method A, use equation (5.3), with  $AR_T(H_0)$  calculated from best fit values given above; method B, use equation (5.4), with experimental values for  $AR_T(H_0)$ . For clarity, we arrange the equations so that all of the experimental data used go on the left of the equations. We treat  $\beta \rho_{Co}^*$  and  $\gamma AR_{Co/Ag}^*$  as two more parameters for fitting the  $H_s$  data with either method, and thereby calculate  $\beta$  and  $\gamma$ .

Method A: Since the best fit values for  $AR(H_0)$  are used, the experimental data at  $H_s$  play a much more important role in the fit compared with method B. Experimental data at  $H_0$  affect this fit only through the  $H_0$  fit. The global fit using method A gives:

$$\beta \rho_{Co}^* = 56 \pm 6$$
$$\gamma A R_{Co/Ag}^* = 0.47 \pm 0.02$$

 $\chi^2$  was 231.9 for degrees of freedom 32, and the two quantities are strongly correlated.

,
See appendix D for discussions about  $\chi^2$  contours. Thus we have:

$$\beta = 0.56 \pm 0.07$$
  
 $\gamma = 0.79 \pm 0.04$ 

•

•

•

,

Since  $\rho_F^* = \rho_F/(1-\beta^2)$  and  $R_{F/N}^* = R_{F/N}/(1-\gamma^2)$ , we find the best values for the resistivity of Co and the AR for Co/Ag interfaces:

$$\rho_{Co} = 68 \pm 20 \,\mathrm{n\Omega m}$$
$$AR_{Co/Ag} = 0.22 \pm 0.03 \,\mathrm{f\Omega m^2}$$

Another interesting quantity is the resistance ratio between down and up channels,  $\alpha$  (see definitions equations (4.3)-(4.6)):

$$\alpha_{Co} = \frac{\rho_{Co}^{1}}{\rho_{Co}^{1}} = \frac{1+\beta}{1-\beta} = 3.6 \pm 0.8$$
  
$$\alpha_{Co/Ag} = \frac{AR_{Co/Ag}^{1}}{AR_{Co/Ag}^{1}} = \frac{1+\gamma}{1-\gamma} = 8.6 \pm 0.9$$

Method B: Experimental data are used for  $AR(H_0)$ . Both the  $H_0$  data, and the differences between  $H_0$  and  $H_s$  data, play important roles in the fit. The fluctuations of  $H_s$  data are less important compared with method A. The global fit using method B gives:

$$\beta \rho_{Co}^* = 48 \pm 5$$
  
 $\gamma A R_{Co/Ag}^* = 0.50 \pm 0.02$ 

 $\chi^2$  was 222.5 for degrees of freedom 32, and again the two quantities are strongly

correlated. Thus we have:

$$\beta = 0.48 \pm 0.06$$
  
 $\gamma = 0.84 \pm 0.04$ 

The best values for the resistivity of Co and AR for the Co/Ag interfaces are:

$$\rho_{Co} = 77 \pm 12 \,\mathrm{n}\Omega\mathrm{m}$$
$$AR_{Co/Ag} = 0.18 \pm 0.02 \,\mathrm{f}\Omega\mathrm{m}^2$$

The down and up channel resistance ratios,  $\alpha$ , are:

$$\alpha_{Co} = 2.8 \pm 0.7$$
  
 $\alpha_{Co/Ag} = 11.5 \pm 1.0$ 

To see how well these parameters fit our chosen four sets of samples, how much difference methods A and B make, and how well the global fit parameters describe the other three sets of Co/Ag samples not included in the global fit, we replot our data in Figures 5.25-5.31. The lines in the figures are the best fits. The solid lines going through the  $H_s$  data and the  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  data are best fit lines calculated from method A. The dotted lines are calculated from method B. Notice that the solid and dotted lines for the three sets of samples with  $t_{Co} = 6$ nm are almost identical, and that they do not differ much in the other four sets.

The best fit lines for the four sets of samples that were included in the global fit match the data well, except that the data points for smaller Ms in the set of equal thickness samples fall systematically below the best fit lines (see Figure 5.28). This is consistent with our finding from other sets of samples with varying Co thickness;



Figure 5.25. Global fit of  $[Co(6nm)/Ag(6nm)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.26. Global fit result of  $[Co(6nm)/Ag(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.27. Global fit result of  $[Co(2nm)/Ag(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.28. Global fit result of  $[Co(t)=Ag(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1 All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.29. Global fit result of  $[Co(t)/Ag(6nm)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.30. Global fit of  $[Co(6nm)/Ag(t)] \times 60$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.31. Global fit result of  $[Co(t)/Ag(6nm)] \times 60$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure 5.32. Global fit of  $[Co(6nm)/AgSn(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $AR_T(H_0) - AR_T(H_s)$ , and (c)  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M - 1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b), (c) are calculated from method (A) and dotted lines are from method (B); see text for details.

that is, our data tend to be smaller than predicted for samples with Co thickness larger than 12nm (section 5.4.6, 5.4.8). We attribute the smaller resistance in thicker Co multilayers to different magnetic structures. Detailed studies of the magnetic structure dependence on Co layer thickness are needed to check this attribution.

By examining the down and up channel resistance ratios,  $\alpha$ 's, for bulk Co and for Co/Ag interfaces, we can ascertain the relative importance of bulk and interface contributions to giant MR. The fits give  $\alpha$ s for Co/Ag interfaces 2 to 4 times larger than those for bulk Co. Thus, we find that the interface contribution is more important to the MR in our sputter-deposited Co/Ag multilayers.

In references [163, 164] we fixed  $2AR_{Nb/Co}$  at independently measured  $6 f\Omega m^2$  and had a global fit results for three sets of sample mentioned earlier ( $H_0$  fit):

$$2AR_{Nb/Co} = 6 \text{ f}\Omega\text{m}^2 \text{ assumed}$$

$$\rho_{Ag} = 10 \pm 3 \text{ n}\Omega\text{m}$$

$$\rho_{Co}^* = 107 \pm 10 \text{ n}\Omega\text{m}$$

$$\beta = 0.48 \pm 0.05$$

$$\alpha_{Co} = 2.9^{+0.5}_{-0.3}$$

$$AR_{Co/Ag}^* = 0.56 \pm 0.03 \text{ f}\Omega\text{m}^2$$

$$\gamma = 0.85 \pm 0.03$$

$$\alpha_{Co/Ag} = 12^{+5}_{-2}$$

In Table 5.25, we list our global fit values, and compare some of them with independently measured results. They almost all agree to within the specified uncertainties.

If we use the AF state equation in the two channel model to fit our data at  $H_p$ , we can do a similar analysis as in this section. The results are also listed in Table 5.25

Table 5.25. Comparison of independent measurement,  $H_0$  fit, in which  $H_0$  and  $H_s$  CPP AR data are used, and  $H_p$  fit, in which  $H_p$  and  $H_s$  data are used. (A) and (B) are two different ways to fit  $H_s$  data, see text for details. The unit for AR is  $f\Omega m^2$ , for resistivity is  $n\Omega m$ .

|                      | independent   | H <sub>0</sub> fit  |               | $H_p$ fit      |               |
|----------------------|---------------|---------------------|---------------|----------------|---------------|
|                      | measurement   |                     |               |                |               |
|                      | (section 3.2) | (A)                 | (B)           | (A)            | (B)           |
| 2AR <sub>Nb/Co</sub> | $6.1 \pm 0.3$ | $6.9 \pm 0.6$       |               | $7.3 \pm 0.6$  |               |
| $\rho_{Ag}$          | $10 \pm 2$    | $7.3 \pm 1.9$       |               | $10.9 \pm 1.9$ |               |
| ρ <sub>Co</sub>      | $68 \pm 10$   | $68 \pm 20$ 77 ± 12 |               | $70 \pm 31$    | $77 \pm 34$   |
| $\rho_{Co}^*$        |               | $100 \pm 6$         |               | $84 \pm 6$     |               |
| $\boldsymbol{\beta}$ |               | $0.56\pm0.07$       | $0.48\pm0.06$ | $0.41\pm0.08$  | $0.29\pm0.06$ |
| $\alpha_{Co}$        |               | $3.6\pm0.8$         | $2.8\pm0.7$   | $2.4\pm0.9$    | $1.8\pm0.7$   |
| $AR_{Co/Ag}$         |               | $0.22\pm0.03$       | $0.18\pm0.02$ | $0.19\pm0.03$  | $0.15\pm0.02$ |
| $AR^{*}_{Co/Ag}$     |               | $0.60\pm0.02$       |               | 0.45 =         | ± 0.02        |
| $\gamma$             |               | $0.79\pm0.04$       | $0.84\pm0.04$ | $0.77\pm0.05$  | $0.82\pm0.05$ |
| $\alpha_{Co/Ag}$     |               | $8.6\pm0.9$         | $11.5\pm1.0$  | $7.5\pm1.0$    | $10.1\pm1.2$  |

(see appendix E for details and discussion).

#### 5.5.1 Co/AgSn samples

We assume that the presence of 4at.% Sn does not change the interface resistance  $R_{Co/Ag}$  much. Since at  $H_0$ ,  $AR_T$  is linear in the resistivity of AgSn, by using the best fit values at  $H_0$  for  $AR_{Nb/Co}$ ,  $\rho_{Co}^*$ , and  $AR_{Co/Ag}^*$ , we can fit the Co/AgSn samples to get the resistivity for AgSn alloy.

A weighted least square fit gives:

$$\rho_{AgSn} = 185 \pm 11 \,\mathrm{n\Omega m}$$

,

 $\chi^2$  was 11.5 for degrees of freedom 5. This is in reasonable agreement with the independently measured value of  $221 \pm 36 n\Omega m$  (see section 3.2). Figure 5.32 shows our data and the best fit lines. The lines in part (a) turn downward at smaller M

because we assumed that the outermost AgSn layer has zero resistance, and the AgSn layer thickness is thicker at smaller M. The close match between the best fit line and our data for  $AR_T(H_0) - AR_T(H_s)$  -- see part (b) of the figure -- is strong evidence for the applicability of the two channel model.

#### 5.6 Summary

In this chapter, we first analyzed our CPP and CIP MR data qualitatively. We showed that the CPP MR is systematically larger than the CIP MR in our Co/Ag and Co/AgSn samples. By inspecting the MR dependence on different layer thicknesses, and comparing with Co/Cu samples, we found evidence for interlayer ferromagnetic ordering between Co layers when the interlayer is 4nm or less. No evidence for such ordering is found for interlayer thicknesses equal to or larger than 6nm. Thus, for our quantitative analysis, we limit ourselves to samples with interlayer at least 6nm thick.

The CPP AR for various single sets of samples having interlayers at least 6nm thick, are consistent with form of the two channel series resistance model. The worst agreement occurs for samples with thick Co layers; we attribute these deviations to internal magnetic structures in the layers. Due to the small amounts of data for each fit, the coefficients sometimes varied a lot. We thus did a global fit to a selected sets of samples to derive all of the unknowns in the two channel model. Comparing our fitting results with independently measured values where available, we found agreement to within mutual uncertainties. Other unknowns, such as the interfacial resistance times area for Co/Ag, the resistance ratios between down and up electron channels for Co/Ag interfaces and for bulk Co, were also derived. From these numbers, we conclude that the interfacial contribution to the MR is important.

## **CHAPTER 6**

### **Summary and Conclusions**

In this dissertation we described the first measurements of the giant magnetoresistance (MR) with current flowing perpendicular to the layer planes (CPP). In all prior studies of the giant MR in magnetic multilayers, the resistance was measured with current flowing in the multilayer planes (CIP). We showed both how to measure the MR in CPP geometry and how to analyze the resulting data. To ensure uniform current through the multilayer, we use superconducting leads. To measure the small resistance associated with the wide and thin geometry of the multilayer, we use a combination of a SQUID and a high precision current comparator. The advantages of our measuring technique are the uniform measuring current, and that the sample geometry is well defined. The disadvantage is that we are limited to low temperature.

By a dc magnetron sputter deposition system, we fabricated Co/Ag and Co/AgSn multilayers on which the CPP and CIP MR can be measured simultaneously. A new masking system was designed for this purpose. The multilayer samples were characterized by X-ray diffraction to confirm the crystalline structures and the desired bilayer thicknesses. A Dektak surface profiler was used to measure the widths of the Nb leads, from which we calculated the effective CPP sample areas.

We modified a cryostat so that we can apply magnetic fields to our samples. A magnetic shield was added outside of our magnet and an electrical noise filter was

put in the existing SQUID circuit to make the measurement feasible.

When we applied magnetic fields to our samples, there were three resistance states noticeable on a resistance versus field plot:  $R(H_0)$ ,  $R(H_s)$ , and  $R(H_p)$  (see figure 3.2). At  $H_s$ , the magnetizations of all the Co layers are aligned with the external field. The magnetic structures of our samples at  $H_0$  and  $H_p$  are poorly understood; detailed studies are not yet available. Almost all of our samples had larger resistances at  $H_0$ than at  $H_p$ . For each sample, the resistance at  $H_p$  is retrievable when cycling the field, but the resistance at  $H_0$  is not. On the other hand, when we made identical samples, resistances at all three fields are reproducible. The reproducibilities for nominally identical samples were checked. The fluctuation of the effective CPP sample area times the total resistance, AR, at selected magnetic fields is about 5% of the average values; similarly, the fluctuation is 10% for CPP MR and 20% for CIP MR.

We found that the CPP MR is larger than the CIP MR for all our samples.  $H_p$  for the CPP and CIP MRs are similar, both slightly larger than the coercive field.

We studied the MR dependence on different Co and/or Ag (AgSn) thicknesses systematically. When the Ag (AgSn) interlayer is 2nm, the MR is very small. We attribute this to interlayer ferromagnetic ordering between adjacent Co layers. For 4nm thick interlayers, there might be some partial ordering. When the interlayer is 6nm or thicker, there is no evidence of such ordering. When the Co thickness was changed, the resistances at  $H_0$  and  $H_p$  showed evidence that the magnetic structures start to change for 12nm Co and larger. For samples with 30nm and 50nm Co, both MR and magnetization curves showed clear shoulders and steps.

The advantage of the CPP geometry over the CIP geometry is that we can analyze our CPP data utilizing a simple two channel series resistance model and thereby separate out bulk from interface scattering. In this model, spin up and spin down conduction electrons are treated in two separate channels, i.e. the spin diffusion length is treated as infinite. The resistance for each spin channel is obtained by simply treating each layer and interface as individual resistances in series. We used the ferromagnetic alignment equation in the model for our  $H_s$  data and associated the antiferromagnetic alignment equation with  $H_0$  data in one case and with  $H_p$  data in the other. Fitting our data with Ag (AgSn) thicknesses 6nm and larger, and Co thicknesses 20nm and smaller, we found that when the Co layer thickness was kept fixed in a set of samples, the model generally had good agreement with our data. When the Co thicknesses were varied, the model did not describe our data as well.

We then concentrated on four sets of data to do a global fit. Three sets were with Co thickness kept fixed, and the fourth with Co and Ag thicknesses kept equal to each other. The fitting determined all parameters in the two channel model. The resulting Co and Ag resistivities, and the Nb/Co interface resistance, agreed with independently measured numbers to within mutual uncertainties. The resulting resistance ratios (between up and down channels) for the interface and bulk Co are strongly correlated. The best values we got showed that the interface contribution to the giant MR effect is very important.

To see how much our results would change with different ways of fitting, we compared the global fit results above with another global fit, where only three sets of data were used and the Nb/Co interface resistance was taken as the independently measured value. The two different global fit results agreed within mutual uncertainties.

The superiority of the two channel model over other models in explaining our CPP data shows clearly in Figure 5.24. The resistivities of our pure Ag and AgSn alloy differ by about a factor of 20. When we plot the quantity  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  versus M-1 for our Co/Ag and Co/AgSn samples, the data fall closely on one straight line. In the two channel model,  $\sqrt{[AR_T(H_0) - AR_T(H_s)]AR_T(H_0)}$  is independent of the resistivity of the spacer layer, just what we find. This is important support for the two channel model.

Now that we have shown how to measure and analyze the CPP MR, we expect it

to become an essential complement to the CIP MR for understanding giant MR. We have shown how to isolate and get quantitative numbers for the interface resistances. Our results showed more directly than previously the importance of the interfaces to the giant MR.

For future work, we have set up two other measuring systems so that high magnetic fields (up to 4 tesla) are available. High fields are needed to study strongly coupled multilayers, like Fe/Cr. Work already in progress includes: (1) studies of other systems such as Co/Cu, Fe/Cr, and Permalloy/Cu, etc. (2) effects of reduced spin diffusion length, studied by adding impurities that flip spins in the non-magnetic spacer layer. (3) normal metal/normal metal multilayers. (4) spin valve structures, where one can control the magnetization orientations of different magnetic layers by applied fields.

# **APPENDICES**

### APPENDIX A

# Current Uniformity in CPP Geometry

#### A.1 Current Uniformity

We analyze the detailed current distribution in our CPP measuring technique in this section. The analysis was made qualitatively by J. Bass, W. P. Pratt Jr., and J. Hetherington. The detailed calculations were made by J. Hetherington.

Consider two long, thin superconducting strips of width W that are in a "crossed" arrangement on opposite sides of a slab of thickness t, diameter 2R, and conductivity  $\sigma$ , see Figure A.1. The potential difference between the strips and the total current flowing between them are V and I, respectively. For our samples, we have  $W \approx 1$ mm,  $2R \approx 4$ mm, and  $t \approx 1\mu$ m. A perspective view of this geometry is also shown in the figure, where the slab is omitted for clarity and the dashed lines indicate the boundary of the overlap region  $W \times W$ . The " edge"  $(I_e)$  and "corner"  $(I_c)$  fringing currents are indicated for one side of the top strip.

#### Summary of Mathematical Analysis.

Edge currents  $I_e$  flows from near an edge of one strip to the opposite strip. These currents are significant only within width t of the  $W \times W$  boundary. With four



Figure A.1. Top and perspective view of crossed-strips geometry.

edges,  $I_e = (VW^2\sigma/t)4\alpha(t/W)$ , where  $\alpha = (\ln 4)/\pi$  see section A.1.1.

Corner currents  $I_c$  flows between the corner-shared extended edges (beyond  $W \times W$ ), as shown. The E-field will vary as 1/r where r is the radial distance from a corner. Since at the four corners the shortest distance between the strips is t, we integrate from r = t to r = R, and obtain  $I_c \approx (VW^2\sigma/t)4\beta(t/W)^2\ln(R/t)$ , where  $\beta = 2/\pi$ , see section A.1.2.

The total current should thus be

$$I = \frac{VW^2\sigma}{t} \left[1 + 4\alpha(\frac{t}{W}) + 4\beta(\frac{t}{W})^2 \ln(\frac{2W}{t}) + \text{HigherOrder}\right] \quad ; \qquad (A.1)$$

and in terms of the correct AR(CPP), we obtain

$$AR(CPP) = \frac{t}{\sigma}$$
(A.2)  
=  $(\frac{W^2V}{I})[1 + 4\alpha(\frac{t}{W}) + 4\beta(\frac{t}{W})^2\ln(\frac{2W}{t}) + \text{HigherOrder}]$  (A.3)

With  $W = 10^3 t$ , AR(CPP) differs from  $t/\sigma$  by less than 1%.



Figure A.2. (a) Evaluate  $\alpha$  for  $I_e$ ; (b) evaluate  $\beta$  for  $I_c$ .

#### A.1.1 Evaluate $\alpha$ for $I_e$

In Figure A.2(a), the y = t and y = 0 planes contain the upper and lower strips of Figure A.1, respectively. The third conductor in the y = 2t plane is a mathematical device included to satisfy the boundary conditions imposed by the slab having a free surface in the plane of the upper strip in Figure A.1. In Figure A.2(a), the whole region between y = 0 and y = 2t is filled with material of conductivity  $\sigma$ . The problem is solved by the conformal mapping  $x + iy = \ln[\cosh(u + iv)]$ , where the lines of constant v in the x,y plane are equipotentials.

Upon scaling we obtain the potential  $\Phi = \text{Im}(2V/\pi)\cosh^{-1}[\exp \pi (x + iy)/2t]$ . Computing the vertical electric field at the lower conducting slab surface from

$$E = \frac{\partial \Phi}{\partial y}|_{y=0}$$

we find directly that

$$E(x) = \frac{V}{t} [e^{-2\xi} + 1]^{-1/2}$$

,

where  $\xi = \pi x/2t$ . Integrating along the bottom plate in Figure A.2(a) from  $x = -\infty$ 

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to +W, we obtain

$$I_{e} = 4W\sigma \int_{-\infty}^{0} E(x) \, dx + 4W\sigma \int_{0}^{W} [E(x) - \frac{V}{t}] \, dx \quad . \tag{A.4}$$

Note for equation A.4 that  $I_e$  is defined in terms of the *deviation* of E(x) from the ideal case where E(x) = 0 for  $-\infty < x < 0$  and E(x) = V/t for 0 < x < W. In the limit  $e^{-W/t} \ll 1$ , we find that  $\alpha = \ln 4/\pi$ .

#### A.1.2 Evaluate $\beta$ for $I_c$

Figure A.2(b) shows how we approximate the 3D problem (for the upper righthand corner of Figure A.1) by a 2D geometry, where the inner cutoff is t, the minimum distance between the two strip extensions in Figure A.1. Thus we are ignoring the component of the field lines that is perpendicular to the slab. We use the conformal mapping  $x + iy = \exp(u + iv)$ . Now the electric field falls off as 1/r from the center, which yields a logarithmic term  $E = (2/\pi)(V/r)$ ,

$$I_c = 4\sigma t \int_t^R E(r) dr = \frac{\sigma t 8V}{\pi} \ln \frac{R}{t} \quad . \tag{A.5}$$

Therefore,  $\beta = 2/\pi$ .

## **APPENDIX B**

# Total Sample Thickness and X-ray Results

On some of our samples, we tried to determine the total sample thickness carefully by dektak surface profiler. These results are listed in table B.1. The uncertainties are estimated to be 3%.

Bilayer thicknesses derived from x-ray spectra of our samples, on which we measure the CPP and CIP MRs simultaneously, are listed here in the order of sample numbers. All samples are on sapphire substrates except samples in run 330, in which Si substrates were used. Computer programs in Quick Basic written by M. Wilson are available to fit x-ray spectra to get the bilayer thicknesses. All results listed here are derived from high angle scans.

Table B.1. Total sample thickness from dektak results of some  $[Co/Ag] \times bilayer-$ number samples, all with nominal total thickness 720nm. Thicknesses are in nm.

| sample | $[Co/Ag] \times M$  | dektak |  |
|--------|---------------------|--------|--|
| no.    |                     | result |  |
| 246-01 | [6/6]×60            | 695    |  |
| 246-05 | $[8/8] \times 45$   | 701    |  |
| 246-09 | $[15/15] \times 24$ | 692    |  |
| 251-01 | $[6/2] \times 90$   | 708    |  |
| 251-03 | $[6/4] \times 72$   | 701    |  |
| 251-04 | $[6/9] \times 48$   | 701    |  |
| 251-05 | $[6/12] \times 40$  | 714    |  |
| 252-01 | [6/12]×40           | 734    |  |

| sample | $[Co/Ag] \times M$  | x-ray          |
|--------|---------------------|----------------|
| no.    |                     | result         |
| 246-01 | $[6/6] \times 60$   | $11.4 \pm 0.1$ |
| 246-05 | $[8/8] \times 45$   | $15.2 \pm 0.1$ |
| 246-09 | $[15/15] \times 24$ | ?              |
| 251-01 | [6/2]×90            | $7.7 \pm 0.1$  |
| 251-03 | $[6/4] \times 72$   | $9.6 \pm 0.1$  |
| 251-04 | $[6/9] \times 48$   | $14.3 \pm 0.1$ |
| 251-05 | [6/12]×40           | $17.2 \pm 0.1$ |
| 252-01 | [6/12]×40           | $17.0 \pm 0.1$ |
| 252-02 | [6/18]×30           | $22.2 \pm 0.1$ |
| 252-03 | $[6/22] \times 26$  | ?              |
| 256-01 | $[6/15] \times 34$  | $21.4 \pm 0.2$ |
| 256-02 | [6/60]×11           | ?              |
| 256-03 | $[6/15] \times 34$  | ?              |
| 256-05 | $[6/35] \times 18$  | ?              |
| 260-01 | [20/20]×18          | ?              |
| 260-02 | $[30/30] \times 12$ | ?              |
| 260-03 | $[50/50] \times 7$  | ?              |
| 261-01 | $[2/6] \times 90$   | 8.3±0.1        |
| 261-02 | $[2/8] \times 72$   | $10.5 \pm 0.1$ |
| 261-03 | $[2/12] \times 51$  | $14.1 \pm 0.1$ |
| 261-04 | $[2/20] \times 33$  | $22.4 \pm 0.1$ |
| 261-05 | $[2/58] \times 12$  | ?              |
| 276-01 | $[1.5/6] \times 96$ | $7.2 \pm 0.1$  |
| 276-02 | $[4/6] \times 72$   | $9.1 \pm 0.2$  |
| 276-03 | $[12/6] \times 40$  | $16.2 \pm 0.2$ |
| 276-04 | [60/6]×11           | ?              |
| 297-01 | $[2/6] \times 60$   | 7.8±0.1        |
| 297-03 | [9/6]×60            | $15.2 \pm 0.1$ |
| 298-01 | [6/6]×60            | $11.6 \pm 0.1$ |
| 298-03 | [20/6]×60           | ?              |
| 300-01 | [6/4]×60            | $9.9 \pm 0.1$  |
| 300-03 | $[6/12] \times 60$  | $17.2 \pm 0.1$ |
| 308-01 | [15/6]×60           | ?              |
| 308-02 | [6/15]×60           | $19.5 \pm 0.1$ |
| 310-01 | [6/6]×60            | $11.4 \pm 0.1$ |
| 310-03 | $[6/5] \times 60$   | $10.4 \pm 0.1$ |
| 310-04 | [6/9]×60            | $14.4 \pm 0.1$ |
| 330-01 | [6/6]×60            | $10.7 \pm 0.1$ |
| 330-02 | [6/30]×20           | ?              |
| 330-03 | [18/18]×20          | ?              |
| 330-04 | [12/6]×40           | $16.9 \pm 0.2$ |
| 333-01 | 6/30 × 20           | ?              |

ł

| sample | $[Co/Ag] \times M$   | x-ray          |
|--------|----------------------|----------------|
| no.    |                      | result         |
| 333-02 | [6/6]×60             | $11.5 \pm 0.1$ |
| 334-01 | [18/18]×20           | ?              |
| 334-02 | [6/60]×11            | ?              |
| 336-01 | [6/6]×60             | $11.0 \pm 0.1$ |
| 336-02 | $[6/30] \times 20$   | ?              |
| 336-03 | [18/18]×20           | ?              |
| 336-06 | [12/6]×40            | $16.2 \pm 0.1$ |
| 336-07 | [6/60]×11            | ?              |
| 336-08 | $[12/12] \times 30$  | ?              |
| 337-01 | [6/6]×60             | $11.7 \pm 0.1$ |
| 337-02 | $[12/12] \times 30$  | ?              |
| 337-03 | $[6/6] \times 1.5$   | ?              |
| 337-04 | $[6/6] \times 3.5$   | ?              |
| 337-05 | $[6/6] \times 5.5$   | ?              |
| 338-01 | [9/6]×48             | $14.7 \pm 0.1$ |
| 338-02 | $[6/60] \times 11$   | ?              |
| 338-03 | $[12/6] \times 40$   | $16.7 \pm 0.2$ |
| 338-04 | [2/6]×90             | $7.6 \pm 0.1$  |
| 338-05 | [6/4]×60             | $9.7{\pm}0.1$  |
| 338-06 | $[6/20] \times 60$   | ?              |
| 338-07 | [6/12]×60            | $16.9 \pm 0.2$ |
| 339-01 | [6/6]×60             | $11.4 \pm 0.1$ |
| 339-02 | [9/6]×48             | $14.2 \pm 0.1$ |
| 340-01 | [2/6]×90             | $7.8 \pm 0.1$  |
| 340-02 | [30/6]×20            | ?              |
| 350-01 | [6/6]×10.5           | $11.9 \pm 0.1$ |
| 350-02 | $[6/6] \times 20.5$  | $11.6 \pm 0.1$ |
| 350-03 | $[6/6] \times 30.5$  | $11.8 \pm 0.1$ |
| 350-04 | [6/4]×60             | $9.8 \pm 0.2$  |
| 350-07 | [6/15]×60            | $19.9 \pm 0.2$ |
| 350-08 | [6/30]×60            | ?              |
| 352-01 | $[12/6] \times 60$   | $17.6 \pm 0.1$ |
| 352-02 | [18/6]×60            | ?              |
|        | $[Co/AgSn] \times M$ | 1              |
| 262-01 | $[6/2] \times 90$    | 7.7±0.1        |
| 262-02 | [6/4]×72             | $9.6 \pm 0.1$  |
| 262-03 | [6/6]×60             | $11.5 \pm 0.1$ |
| 262-04 | $[6/15] \times 34$   | ?              |
| 262-05 | [6/35]×18            | ?              |
| 277-02 | [6/12]×40            | $17.0 \pm 0.1$ |
| 277-03 | 6/60×11              | ?              |
| 277-04 | [6/6]×60             | $11.8 \pm 0.2$ |

Table B.2. X-ray results of  $[Co/Ag] \times bilayer$ -number samples. Thicknesses are in nm.

Ag -59.9256 50.1076 0.1383 0.5400 -0.5500 4.6000 Co -34.4127 27.9505 0.1830 0.2700 -2.2000 3.8000 0.0000 0.7000 0.0000 0.0000 0.0000 0.0000 2740.0 7200.0 26.0IIHL ICA LAT LOG GEO MOD IXR XR-lamda1 XR-lamda2 WTEXP 0 0 1 0 1 1 0 1.5406 1.5444 1.50NNBI it tol icon cfw 60 2 0.010 1 0.050 9.5220 0.0012 -1.0000 0.2496 24.2637 1.2500 28.1073 1.5000 2.3631 2.0395 -0.0192 0.0431 0.5000 0.5000 -1.0000 -1.0000 0.0000 3.0000 1.0000 1.0000 0.0000 0.0000 0.0000 0.0000 0.0000 1.0000 0.0000 1.0000 1.0000 1.0000 1.0000 1.0000

0100101

1001111

1111111

1111111

SUPREX 8.0

12-MAY-94

variable input file = last\_input.txt

data input file = sucoag.sav

HIGH ANGLE CALCULATION

model # 1

SUPREX8 output file of the fit in Figure 2.2.

Element #1: Ag Element #2: Co CHI SQUARED = 2.770 WT Exp = 1.5000 average x-ray lamda used x-ray lamda = 1.5419

absorbtion thickness = 2740.000 7200.000
monochromator 2 theta = 26.00000
1/sin(2th) used
nlat = 1

na distribution

- 18 0.0004
- 19 0.0115
- 20 0.1813
- 21 1.5538
- 22 7.3361
- 23 19.3316
- 24 28.9268
- 25 25.0882
- 26 12.8533
- 27 3.9266
- 28 0.7110

- 29 0.0749
- 30 0.0045
- 31 0.0001

nb distribution

- 21 0.0006
- 22 0.0104
- 23 0.1112
- 24 0.7705
- 25 3.4845
- 26 10.3073
- 27 20.0029
- 28 25.5629
- 29 21.6138
- 30 12.1607
- 31 4.5834
- 32 1.1652
- 33 0.2009
- 34 0.0236
- 35 0.0019
- 36 0.0001
- ITER = 2
- NBI = 60
- THKA = 57.112 THKB = 57.669
- STND DEV A = 3.206904
- STND DEV B = 3.179757
- LAST ITERATION # = 0

FLAMDA = 0.000000E+00

|                        | thknes  | s d    | int    | d ave |  |  |  |
|------------------------|---------|--------|--------|-------|--|--|--|
| A =                    | 57.11   | 2.21   | 332.   | 3598  |  |  |  |
| B =                    | 57.67   | 2.21   | 332.   | 0458  |  |  |  |
| strai                  | n profi | le la  | yer A  |       |  |  |  |
| -0.0                   | 192 -0. | 0116 · | -0.007 | 1     |  |  |  |
| -0.0                   | 192 -0. | 0116 · | -0.007 | 1     |  |  |  |
| strain profile layer B |         |        |        |       |  |  |  |
| 0.0                    | 431 0.  | 0261   | 0.015  | 9     |  |  |  |
| 0.0                    | 431 0.  | 0261   | 0.015  | 9     |  |  |  |

| 1 Packana |     | 0 5000  | 0 0000 | 0 |
|-----------|-----|---------|--------|---|
| I Dackgro | una | 9.5220  | 0.0000 | U |
| 2 Scale   | Fac | 0.0012  | 0.0000 | 0 |
| 3 d int ( | -1) | -1.0000 | 0.0000 | 1 |
| 4         | с   | 0.2496  | 0.0000 | 0 |
| 5         | N A | 24.2637 | 0.0000 | 0 |
| 6 sigma   | N A | 1.2500  | 0.0000 | 1 |
| 7         | NB  | 28.1073 | 0.0000 | 0 |
| 8 sigma   | NB  | 1.5000  | 0.0000 | 1 |
| 9         | d A | 2.3631  | 0.0000 | 0 |
| 10        | d B | 2.0395  | 0.0000 | 1 |
| 11 del    | dA1 | -0.0192 | 0.0000 | 0 |
| 12 del    | dB1 | 0.0431  | 0.0000 | 0 |
| 13 A      | exp | 0.5000  | 0.0000 | 1 |
| 14 B      | exp | 0.5000  | 0.0000 | 1 |
| 15 del    | dA2 | -1.0000 | 0.0000 | 1 |

| 16 del dB2   | -1.0000 | 0.0000 | 1 |
|--------------|---------|--------|---|
| 17 diffusion | 0.0000  | 0.0000 | 1 |
| 18 diff exp  | 3.0000  | 0.0000 | 1 |
| 19inpln denA | 1.0000  | 0.0000 | 1 |
| 20inpln denB | 1.0000  | 0.0000 | 1 |
| 21           | 0.0000  | 0.0000 | 1 |
| 22           | 0.0000  | 0.0000 | 1 |
| 23           | 0.0000  | 0.0000 | 1 |
| 24           | 0.0000  | 0.0000 | 1 |

### APPENDIX C

### Extra Data

There are some data we did not include in the main text because of various reasons listed below: (a) High purity Ar gas supply line leakage, (b) sputtering gun parts for different materials mixed, (c) High substrate temperature, (d) Si substrates instead of sapphire ones, (e) Ag target penetrated through, (f) different deposition rates used. We put all these Co/Ag data prior to sputtering run 352 in this appendix. All Co/Ag(Sn4%) data of runs 262 and 277 are listed in the main text. No additional samples fall into the categories described above.

Samples produced under conditions (a), (b), and (c) usually have larger resistances than expected, because of contamination in (a) and (b), and potentially different structure in (c). (d) Si substrates were used in run 330, no definite conclusions can be drawn about the effects of different substrates. In condition (e), run 336, the center of the Ag target was all consumed within the run. The target seat, which is made of bronze, underneath was sputtered for the last few samples. The deposition rate was reduced and the Ag layers became Ag-bronze alloy ones. These Co/alloy multilayers have similar resistances to the nominally identical Co/Ag samples. The effect of different deposition rates on the structures and hence the resistances of our samples are not yet studied systematically. Because we were making Co/Ag and Co/Cu multilayers alternately, and simply because the first runs of these different

Table C.1. CPP AR and CIP conductance at  $H_0$ ,  $H_p$ , and  $H_s$  of extra  $[Co/Ag] \times bilayer-number samples$ . Layer thicknesses are in nm. The error in CIP conductance is  $\pm 0.0001\Omega^{-1}$ . See text for the comment column.

| sample | $[Co/Ag] \times M$  | CPP $AR_T$ (f $\Omega$ m <sup>2</sup> ) |       |      | CIP C $(\Omega^{-1})$ |        |        | comment        |                   |
|--------|---------------------|---|-------|------|-----------------------|--------|--------|----------------|-------------------|
| no.    |                     | $H_0$                                   | $H_p$ | H,   | error%                | $H_0$  | $H_p$  | H <sub>s</sub> |                   |
| 256-03 | $[6/15] \times 34$  | 86.4                                    | 77.7  | 57.2 | 6.7%                  | 0.7353 | 0.7611 | 0.8386         | С                 |
| 330-01 | [6/6]×60            | 131.6                                   | 102.5 | 79.2 | 3.3%                  | 0.3675 | 0.3657 | 0.3836         | a,b,d,f           |
| 330-02 | $[6/30] \times 20$  | 47.3                                    | 46.8  | 39.1 | 3.3%                  | 0.9897 | 1.0007 | 1.0347         | b,d,f             |
| 330-03 | [18/18]×20          | 56.0                                    | 54.3  | 48.9 | 2.5%                  | 1.0147 | 1.0275 | 1.0460         | b,d,f             |
| 330-04 | $[12/6] \times 40$  | 93.4                                    | 76.1  | 69.1 | 2.3%                  | 0.4685 | 0.4793 | 0.4950         | b,d,f             |
| 333-01 | $[6/30] \times 20$  | 64.9                                    | 59.6  | 51.9 | 2.8%                  | 1.1694 | 1.1839 | 1.2126         | a,b,f             |
| 333-02 | [6/6]×60            | 111.4                                   | 98.4  | 72.6 | 2.9%                  | 0.4530 | 0.4877 | 0.5270         | b,f               |
| 334-01 | [18/18]×20          | 88.7                                    | 80.3  | 75.1 | 3.7%                  | 0.8461 | 0.8552 | 0.8651         | a,b, <sup>g</sup> |
| 334-02 | [6/60]×11           | 41.8                                    | 39.4  | 36.2 | 1.4%                  | 1.7937 | 1.8038 | 1.8272         | b                 |
| 336-06 | $[12/6] \times 40$  | 72.1                                    | 70.4  | 66.1 | 2.7%                  | 0.4645 | 0.4946 | 0.5118         | e                 |
| 336-07 | [6/60]×11           | 35.7                                    | 33.4  | 30.3 | 3.5%                  | 1.2102 | 1.2181 | 1.2330         | e                 |
| 336-08 | $[12/12] \times 30$ | 83.2                                    | 66.7  | 57.0 | 4.1%                  | 0.5204 | 0.5557 | 0.5836         | e                 |
| 338-01 | $[9/6] \times 48$   | 125.6                                   | 97.5  | 92.3 | 4.0%                  | 0.3972 | 0.4382 | 0.4512         | a                 |

<sup>g</sup>Misaligned CIP sample (see section 2.2).

multilayers were made by different persons, the Co deposition rate used in Co/Ag is higher than the one used in Co/Cu. In some Co/Ag runs, the lower Co deposition rate was used by mistake; in some others the equipment failed to function at the higher deposition rate and a lower rate were used.

All these samples are listed here in the order of sample numbers.

### APPENDIX D

### **Global Fit Details**

For general information about error analysis, please see reference [165]. What follows is based on reference [166].

Assume we have l common unknowns,  $c_1 \cdots c_l$ , in different sets of data. Each set of data can be described by a theoretical relation y = Y(x) where y and x are experimentally measurable quantities. The unknowns  $c_1 \cdots c_l$  appear in the coefficients of the x's. We want to fit these sets of data simultaneously to get the unknowns. Y must be linear in c's to do this global fit the way we describe here.

In set p we have n(p) pairs of data  $(x_i, y_i)$ , each pair is related theoretically by  $y_i = Y(x_i, c_1, \dots, c_l, p)$ . The deviation of the measured  $y_i$  from the theoretical value, weighted by the uncertainty of  $y_i, \sigma_i$ , is then

$$\frac{y_i(p) - Y(x_i, c_1, \cdots, c_l, p)}{\sigma_i(p)}$$

 $\chi^2(p)$  for this set of data is then

$$\chi^{2}(p) = \sum_{i=1}^{n(p)} \frac{[y_{i}(p) - Y(x_{i}, c_{1}, \cdots, c_{l}, p)]^{2}}{\sigma_{i}^{2}(p)}$$

To do a global fit to all sets of data, we minimize

$$\chi^2_{\rm tot} = \sum_p \chi^2(p)$$

•

We limit ourselves to the case that all functions Y are linear in the c's. Thus we can write

$$Y(x_i, c_1, \cdots, c_l, p) = \sum_{k=1}^l c_k \cdot g_k(x_i, p)$$

where  $g_k(x_i, p)$  are functions of  $x_i$  in different sets p. Note that if there is a constant term independent of the c's on the right side, we can move it to the left and redefine the Y's. We have now,

$$\chi_{\text{tot}}^{2} = \sum_{p} \sum_{i=1}^{n(p)} \frac{[y_{i}(p) - Y(x_{i}, c, p)]^{2}}{\sigma_{i}^{2}(p)}$$

where c denotes all l unknowns. To get the best fit values  $c^*$  for all the unknowns, we minimize  $\chi^2_{tot}$  with respect to the c's:

$$\begin{aligned} \frac{\partial \chi^2}{\partial c_j} &= 0 \\ &= -2 \sum_p \sum_{i=1}^{n(p)} \frac{[y_i(p) - Y(x_i, c, p)] \cdot g_j(x_i, p)}{\sigma_i^2(p)} \\ &= -2 \sum_p \sum_{i=1}^{n(p)} \left[ \frac{y_i(p) \cdot g_j(x_i, p)}{\sigma_i^2(p)} - \frac{g_j(x_i, p)}{\sigma_i^2(p)} \cdot \sum_{k=1}^l c_k^* \cdot g_k(x_i, p) \right] \end{aligned}$$

Define the data vector

$$U_j = \sum_p \sum_{i=1}^{n(p)} \frac{y_i(p) \cdot g_j(x_i, p)}{\sigma_i^2(p)}$$

and the symmetric matrix

$$M_{jk} = \sum_{p} \sum_{i=1}^{n(p)} \frac{g_j(x_i, p) \cdot g_k(x_i, p)}{\sigma_i^2(p)}$$

•

We then write

$$U_l = \sum_{k=1}^l M_{jk} \cdot c_k^*$$

,

or simply

$$\mathbf{U} = \mathbf{M} \cdot \mathbf{c}^*$$

The best fit values for the unknowns are solved by

$$\mathbf{c}^* = \mathbf{M}^{-1} \cdot \mathbf{U}$$

 $M^{-1}$  is known as the "error matrix". Its diagonal elements are the squares of the errors,  $(\sigma c_k)^2$ , and the off-diagonal elements are the covariances, which are the correlations between the best fit values  $c_k^*$ .

Computer programs in MathCad were written to do the global fit described above. These programs can be modified for different numbers of unknowns and different sets of data.

#### Discussions of $\chi^2$ contours in the $H_s$ fit method A.

The two fitting parameters in the  $H_s$  fit are strongly correlated. Thus the conclusion we drew,  $\gamma$  is larger than  $\beta$ , has to be examined. We look at the contours of constant  $\chi^2$  of the fit. If we draw contours in steps of  $\chi^2_{min}/32$ , where 32 is the degrees of freedom,

$$\chi_n^2 = \chi_{min}^2 + n \frac{\chi_{min}^2}{32}$$

, we found tilted ellipses. See Figure D.1. The straight line in the figure represents  $\beta = \gamma$ , i.e. the up and down resistance ratios are the same for bulk Co and Co/Ag interface. The line is just outside of the n = 7 contour. Thus our conclusion is good. If we consider the errors in the  $H_0$  fit, which propagate into the  $H_s$  fit, our conclusion is weakened. By changing the four parameters of the  $H_0$  results, we can find out whether increasing or decreasing each parameter will make  $\beta$  and  $\gamma$  become closer.



Figure D.1. Plot of constant  $\chi^2$  contours in the  $H_s$  fit method A. The straight line represents  $\beta = \gamma$ .

This is the most dangerous direction that our conclusion will fail. We change each of the four parameters by amounts so that, in the  $H_0$  fit,  $\chi^2 = \chi^2_{min} + \chi^2_{min}/30$  (30 is the degrees of freedom in the  $H_0$  fit). Then we do the  $H_s$  fit again and look at the contours of constant  $\chi^2$  plot. The  $\beta = \gamma$  line now just intercept with n = 4 contour. Then we change the four parameters in the  $H_0$  fit so that  $\chi^2 = \chi^2_{min} + 2\chi^2_{min}/30$ . The  $H_s$  fit using these parameters has a contours of constant  $\chi^2$  plot with  $\beta = \gamma$  line just intercept with n = 3 contour. To see exactly how  $\beta$  and  $\gamma$  are correlated, further study using a non-linear fit is needed.
## APPENDIX E

## Global Fit to Peak Field CPP AR Data

In this appendix, we do a global fit to our CPP AR data at  $H_p$  and  $H_s$  with the AF state and F state equations in the two channel model, respectively.

We chose the same four sets of Co/Ag samples as in section 5.5:

- 1. Section 5.4.2. Fixed Co, Ag thicknesses at 6nm, various M.
- 2. Section 5.4.3. Fixed Co thickness at 6nm, total thickness at 720nm. *Exclude* samples with  $t_{Ag} < 6$ nm.
- 3. Section 5.4.4. Fixed Co thickness at 2nm, total thickness at 720nm.
- 4. Section 5.4.5. Co and Ag thicknesses kept equal but varying, fixed total thickness at 720nm. Exclude two samples  $t_{Co} = 30, 50$ nm.

 $[Co(6nm)/Ag(6nm)] \times 60$  samples can be put in any category where they appear but should not be repeated. Listing samples repeatedly will cause them to be weighed more than the others. We focus on these samples in the four sets to find the best values for the six unknowns in the model. All of the uncertainties in this section have been adjusted so that the reduced  $\chi^2$  is equal to the degree of freedom. The degree of freedom is equal to the number of data points minus the number of unknowns in the fit. We have 34 data points, 4 unknowns in  $H_p$  fit, and 2 unknowns in  $H_s$  fit. Thus the degree of freedom is 30 for  $H_p$  fit and 32 for  $H_s$  fit. Uncertainties have been multiplied by  $\sqrt{30}$  and  $\sqrt{32}$  in  $H_p$  and  $H_s$  fits, respectively.

Assuming that our CPP AR at  $H_p$  can be described by the AF state equation, we have, from the two channel model:

$$AR_{T}(H_{p}) = 2AR_{Nb/Co} + M\rho_{Co}^{*}t_{Co} + (M-1)\rho_{Ag}t_{Ag} + 2(M-1)AR_{Co/Ag}^{*}(E.1)$$

$$AR_{T}(H_{s}) = AR_{T}(H_{p}) - \frac{\left[M\beta\rho_{Co}^{*}t_{Co} + (M-1)\gamma 2AR_{Co/Ag}^{*}\right]^{2}}{AR_{T}H_{p}}(E.2)$$

$$\sqrt{\left[AR_{T}(H_{p}) - AR_{T}(H_{s})\right]AR_{T}(H_{p})} = M\beta\rho_{Co}^{*}t_{Co} + (M-1)\gamma 2AR_{Co/Ag}^{*}.(E.3)$$

At  $H_p$ , there are  $AR_{Nb/Co}$ ,  $\rho_{Co}^*$ ,  $\rho_{Ag}$ , and  $AR_{Co/Ag}^*$  four unknowns. A global fit, with the same principle as weighted least square fit (see appendix D for details), gives:

$$2AR_{Nb/Co} = 7.3 \pm 0.6 \, \text{f}\Omega\text{m}^2$$

$$\rho_{Co}^* = 84 \pm 6 \, \text{n}\Omega\text{m}$$

$$\rho_{Ag} = 10.9 \pm 1.9 \, \text{n}\Omega\text{m}$$

$$AR_{Co/Ag}^* = 0.45 \pm 0.02 \, \text{f}\Omega\text{m}^2$$

 $\chi^2$  was 125.1.

There are two ways to fit the  $H_s$  data: method A, use equation E.2, with  $AR_T(H_p)$  calculated from the best fit values given above; method B, use equation E.3, with experimental values for  $AR_T(H_p)$ . For clarity, we arrange the equations so that all experimental data go in the left of the equations. We treat  $\beta \rho_{Co}^*$  and  $\gamma AR_{Co/Ag}^*$  as two more unknowns when we fit the  $H_s$  data with either method.

Method A: The global fit using method A gives:

$$\beta \rho_{Co}^* = 34 \pm 6$$
  
 $\gamma A R_{Co/Ag}^* = 0.35 \pm 0.02$ 

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 $\chi^2$  was 528.8. Thus we have:

$$\beta = 0.41 \pm 0.08$$
  
 $\gamma = 0.77 \pm 0.05$ 

Since  $\rho_F^* = \rho_F/(1-\beta^2)$  and  $R_{F/N}^* = R_{F/N}/(1-\gamma^2)$ , we find the best values for the resistivity of Co and AR for Co/Ag interface:

$$\rho_{Co} = 70 \pm 31 \,\mathrm{n\Omega m}$$
  
 $AR_{Co/Ag} = 0.19 \pm 0.03 \,\mathrm{f\Omega m^2}$ 

The resistance ratio between down and up channels,  $\alpha$  (see definitions equations (4.3)-(4.6)) is:

$$\alpha_{Co} = \frac{\rho_{Co}^{\downarrow}}{\rho_{Co}^{\uparrow}} = \frac{1+\beta}{1-\beta} = 2.4 \pm 0.9$$
$$\alpha_{Co/Ag} = \frac{AR_{Co/Ag}^{\downarrow}}{AR_{Co/Ag}^{\uparrow}} = \frac{1+\gamma}{1-\gamma} = 7.5 \pm 1.0$$

Method B: The global fit using method B gives:

$$\beta \rho_{Co}^* = 24 \pm 5$$
  
 $\gamma A R_{Co/Ag}^* = 0.37 \pm 0.02$ 

 $\chi^2$  was 395.5. Thus we have:

$$\beta = 0.29 \pm 0.06$$
  
 $\gamma = 0.82 \pm 0.05$ 

The best values for the resistivity of Co and AR for Co/Ag interface are:

$$\rho_{Co} = 77 \pm 34 \,\mathrm{n}\Omega\mathrm{m}$$
$$AR_{Co/Ag} = 0.15 \pm 0.02 \,\mathrm{f}\Omega\mathrm{m}^2$$

The resistance ratio between down and up channels,  $\alpha$ , is:

$$\alpha_{Co} = 1.8 \pm 0.7$$
  
 $\alpha_{Co/Ag} = 10.1 \pm 1.2$ 

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From our independent measurements, we have (see section 3.2):

$$2AR_{Nb/Co} = 6.1 \pm 0.3 \,\mathrm{f}\Omega\mathrm{m}^2$$
$$\rho_{Co} = 68 \pm 10 \,\mathrm{n}\Omega\mathrm{m}$$
$$\rho_{Ag} = 10 \pm 2 \,\mathrm{n}\Omega\mathrm{m} \quad .$$

Our global fits to the  $H_p$  and  $H_s$  data give good agreement with the resistivities of Co and Ag, but a larger value for the Nb/Co interfaces than independently measured.

In Figures E.1–E.8, we replot our CPP data and the lines show our best fits. The solid lines going through the saturation data and the  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  data are best fit lines calculated from method A; the dotted lines are calculated from method B. As in the case of the  $H_0$  data fits, they do not differ much. Other features we found in the global fit to  $H_0$  data are mostly applicable here, except that the fit to the  $[Co(6nm)/Ag(t)] \times M H_p$  data is not good, see Figure E.2. From that figure, it seems that the M = 60 samples should be excluded from the fit, and maybe even the M = 48 sample if we require straight line behavior. If we exclude the six  $[Co(6nm/Ag(6nm)] \times 60$  samples from the fit, we also have to exclude the whole set of samples Co(6nm)/Ag(6nm) with various bilayer numbers. With only three sets of samples for the  $H_p$  global fit, we can not determine four unknowns uniquely. We have to assume that the independently measured Nb/Co interface resistance is correct and then fit the three unknowns: the Co and Ag resistivities, and the Co/Ag interface resistance.



Figure E.1. Global fit of  $[Co(6nm)/Ag(6nm)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M-1. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.2. Global fit of  $[Co(6nm)/Ag(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.3. Global fit of  $[Co(2nm)/Ag(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.4. Global fit of  $[Co(t)=Ag(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.5. Global fit of  $[Co(t)/Ag(6nm)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M-1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.6. Global fit of  $[Co(6nm)/Ag(t)] \times 60$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M - 1. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.7. Global fit of  $[Co(t)/Ag(6nm)] \times 60$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M - 1. Solid lines for  $H_s$  in (a) and in (b) are calculated from method (A) and dotted lines are from method (B); see text for details.



Figure E.8. Global fit of  $[Co(6nm)/AgSn(t)] \times M$  samples for (a) Area times CPP resistances at  $H_0$ ,  $H_p$ , and  $H_s$  and (b)  $AR_T(H_p) - AR_T(H_s)$ , and (c)  $\sqrt{[AR_T(H_p) - AR_T(H_s)]AR_T(H_p)}$  versus M - 1. All samples have total thickness 720nm. Solid lines for  $H_s$  in (a) and in (b), (c) are calculated from method (A) and dotted lines are from method (B); see text for details.

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