

X-ray Diffraction Studies of Materials for Magnetic Recording Technology

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In magnetic recording technology, the equivalent of Moore's law has the areal density (number of bits per unit area on the recording medium surface) increasing at 60-100% per year, and, of equal interest to the consumer, the cost per megabyte decreasing at 40-50% per year [1]. To achieve this phenomenal progress, many aspects of the magnetic recording system (see Figure 1) must be continuously improved. Of interest to materials scientists are the materials in the read and write heads as well as the recording disk. Here we will describe results obtained with x-ray diffraction at NSLS beamline X20 on thin films used in the read head and the magnetic media and show how these results have provided important understanding into the magnetic properties of the materials used in these devices.

Present recording media consist of several layers sputter deposited onto a glass, or NiP-coated Al, substrate. A typical example is shown in Figure 1(b). The seed layer(s) and Cr underlayer are chosen to produce small grains and to give the desired crystallographic texture. The magnetic recording layer is typically a CoPtCrB alloy with an hcp structure and is a granular film (i.e., comprised of magnetic grains separated by a nonmagnetic phase) with an average grain size, at present, of about 9nm [2, 3]. However, the grain size is decreasing as the areal density increases in an attempt to keep the number of grains per bit constant. If the magnetic grains are too small, the media will be thermally unstable or superparamagnetic with the consequent, rather undesirable, loss of recorded data [1, 4]. The stored magnetic energy is $K_U V$, where V is the grain volume and K_U is the magnetocrystalline anisotropy, and this must be significantly larger than $k_B T$ for thermal stability. To compensate for the

drop in V , research has focused on increasing K_U [4]. One route to thermally stable media is the use of CoPtCrB alloys with high Pt [5], since Pt increases the magnetic anisotropy. However, at Pt concentrations $>30\%$ K_U plateaus, and then drops near 40% Pt. To understand the cause of this drop, we have conducted X-ray scattering studies on these high Pt media.

The thin film media were CoPtCrB alloys with a (10.0) texture and varying Co and Pt concentrations. These were sputter deposited onto Cr/NiAl underlayers with either glass, or NiP coated Al, substrates and were covered with a thin amorphous carbon protective layer. The diffraction measurements were performed at

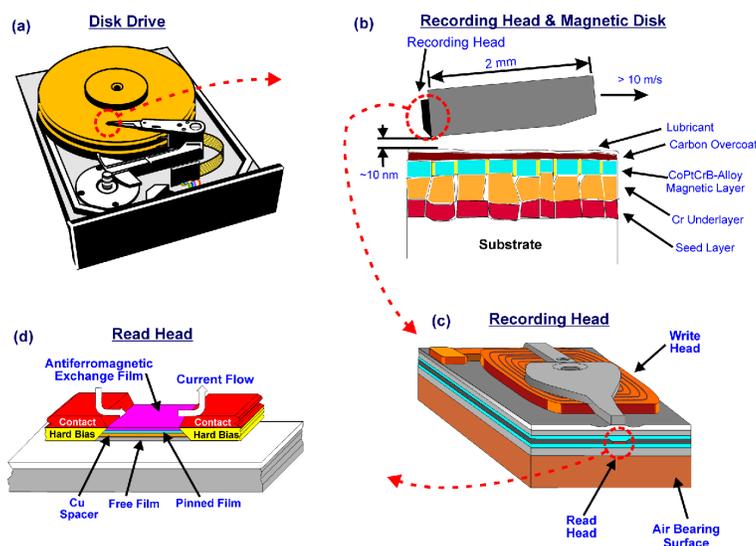


Figure 1. (a) Inside of a typical disk drive. (b) Illustration of the recording head (top) and magnetic disk. In the disk, typical thicknesses for the seed and underlayers are 10-30nm, while the magnetic layer is about 10nm thick. The protective carbon film is about 4-5nm in thickness. There is an air bearing between the head and disk. (c) Schematic of the recording head. This perspective view is approximately 90deg from that in (b). (d) Illustration of the spin valve read head.

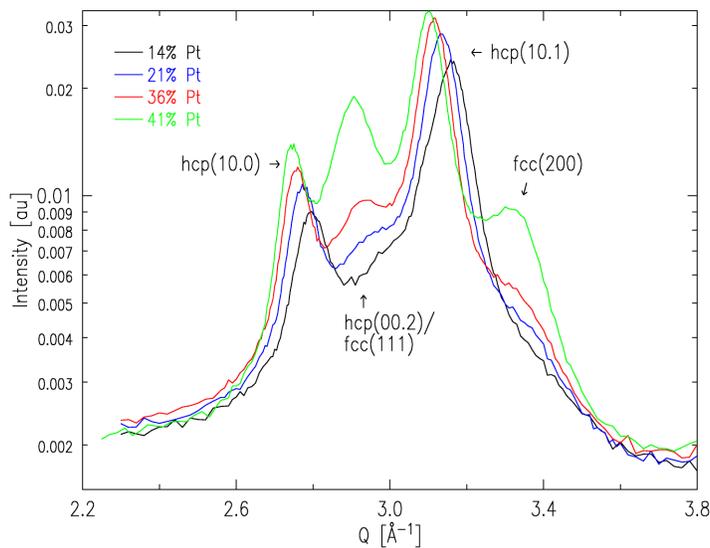


Figure 2. Diffraction pattern from magnetic recording disks with varying Pt concentration in the media. Here Q is the scattering vector with magnitude $Q=(4\pi/\lambda) \sin \theta$, where θ is half the scattering angle. This vector has a (fixed) direction 60deg from the sample normal.

beamline X20C with an energy of 10.3keV using a Ge detector to eliminate the Co and Cr fluorescence. A variety of data were collected to determine the lattice parameters, extent of the (10.0) texture and grain size. Particular attention was given to stacking faults and the presence of any fcc material in the otherwise hcp alloy; using data obtained at NSLS, Dova et al. [6] have shown that these have an important influence on the magnetic properties. Furthermore, the anisotropy of fcc Co is significantly smaller than for hcp Co. Figure 2 shows radial diffraction scans taken with a grazing incidence angle, which limits the diffraction to the top, magnetic layer of interest, and an exit angle of 30deg. For this (10.0) textured media, this scan passes through the maximum in the hcp(10.1) and fcc(200) peaks. Other peaks are apparent since the media is not strongly textured; the mosaic spread is about 25deg. The various peaks are labeled; note that because the peaks are broad, it is impossible to distinguish the fcc(111) and hcp(00.2) peaks. The shifts in the peak positions are due to Pt substituting for Co, which increases the lattice parameters (Pt is much larger than Co or Cr). Most important, the intensity of the fcc(200) and (111) peaks increase significantly with increasing Pt concentration. This results from the formation of an fcc Co-alloy along with the hcp Co-alloy and is reasonable, since Pt forms an fcc structure, while Co is hcp. By profile fitting, we have quantified the amount of fcc phase in the otherwise hcp magnetic alloy. These results show that the formation of this fcc Co-alloy

is the reason the anisotropy plateaus between 28-36% Pt and that the drop in anisotropy near 40% Pt results from a significant increase in the amount of fcc phase at this composition (see Fig. 2). It is rather interesting that the stacking fault density does not increase with Pt concentration, as the formation of such faults is a common mechanism for the hcp-to-fcc transformation [7, 8]. We are presently investigating methods to retard the fcc phase formation.

We now discuss some aspects of the spin valve or giant magnetoresistance (GMR) read head. Figure 1(d) shows the structure of these devices, although in a more simplified form than currently implemented. The antiferromagnetic (AF), exchange biasing layer pins the direction of the reference (pinned) ferromagnetic layer, while the free, ferromagnetic layer rotates due to the magnetic field from the disk. This rotation changes the resistance of the device. Since the total spin valve thickness must significantly decrease for future heads and since the AF film is by far the thickest layer, reducing this thick-

ness is crucially important. However, a critical thickness of the AF pinning layer is needed to develop the exchange pinning [9]. Presently, MnPt is used as an AF layer. This has a chemically disordered fcc structure as deposited, but upon annealing to >240degC, MnPt forms a chemically ordered, $L1_0$ structure, where

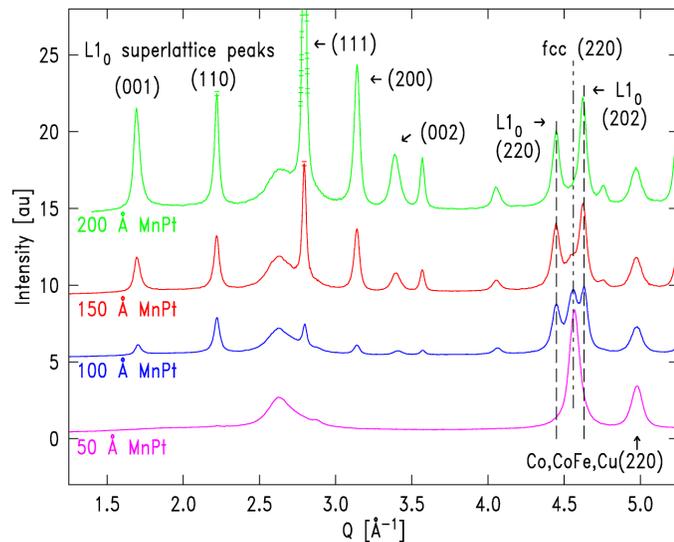


Figure 3. Grazing incidence diffraction pattern for NiFe/CoFe/Cu/Co/MnPt/Ta spin valves as a function of MnPt film thickness. The intensities have been normalized so that the integrated intensity in the fcc(220) and $L1_0$ (202) and (222) peaks are the same. Most of the peaks are labeled; any unlabeled peaks are from MnPt and are unlabeled for clarity. Again Q is the scattering vector (see Fig. 2).

atomic planes along the c-axis are alternately all Mn and all Pt. In this phase, MnPt is antiferromagnetic, while in the fcc phase, MnPt is nonmagnetic.

To better understand the critical thickness in MnPt spin valves, we have conducted x-ray diffraction measurements of the structure and chemical order in annealed MnPt spin valves (substrate/NiFe/CoFe/Cu/Co/MnPt/Ta) as a function of MnPt film thickness (5-25 nm). Figure 3 shows spectra obtained with grazing incidence and exit angles (measuring planes perpendicular to the substrate). There are several structural changes with MnPt thickness. First, the thinnest film is chemically disordered (fcc). However, with increasing thickness, the $L1_0$ superlattice peaks appear signaling the presence of chemical order. Second, focusing on the region of Fig. 3 near $Q=4.5\text{\AA}^{-1}$, the presence of the fcc(220) and the $L1_0$ (202) and (222) diffraction peaks for 10 and 15nm MnPt shows in these films there is two phase coexistence between fcc and $L1_0$ MnPt phases. The extent of chemical order has been quantified using the Warren long-range order parameter, S , which is zero for no chemical order, unity for complete order and is proportional to the number of Pt (or Mn) atoms on the "right" lattice sites [7, 8, 10]. Figure 4 shows that both exchange pinning (H_p) and S develop with increasing thickness and the data show a strong correlation between these quantities. These results show that there is no exchange for thin films because there is no chemical order in these films (i.e., the MnPt

is not antiferromagnetic). They further show that the strong exchange coupling only develops for fully ordered MnPt.

As is also apparent in Fig. 3, the texture changes with increasing thickness as seen by the increasing intensity of the MnPt(111) peak with increasing MnPt thickness. Pole figures of the MnPt(111) peaks (not shown) demonstrate that there is strong (111) texture in the as-deposited and 5nm annealed films. With increasing chemical order, this changes to (111) texture plus a nearly isotropic contribution, and the isotropic part increases with more developed chemical order. It is likely that this change is a result of the nucleation and growth process for the formation of the $L1_0$ phase.

It is interesting to note that NiMn spin valves and FePt films (which is a high K_U material that may be useful as an advanced recording media [4]) show a similar dependence of the chemical order on film thickness. This suggests a common mechanism preventing (or slowing) ordering for thin films. In these annealed films, the chemically ordered phase forms by nucleation and growth as evidenced by the two-phase coexistence mentioned above. It is reasonable to believe that the transformation rate into the $L1_0$ phase is limited by nucleation, since the grains in the $L1_0$ phase are large. If this is the case, then the lack of chemical order in the thin films is due to a lack of nucleation sites, since for a given concentration of nucleation sites, thinner films will have fewer sites.

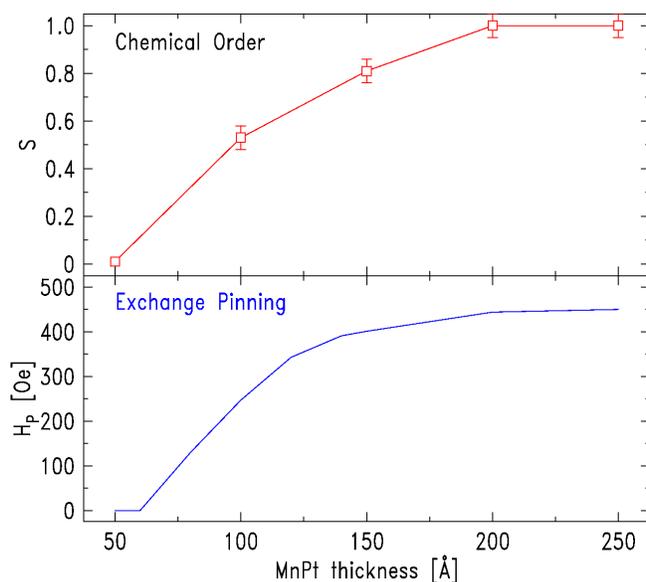


Figure 4. Chemical order (S) and exchange pinning field (H_p) vs MnPt thickness. H_p measures the strength of the magnetic coupling between the antiferromagnetic MnPt and the ferromagnetic Co layers.

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