MAGNETIC MEDIA PERFORMANCE: CONTROL METHODS FOR CRYSTALLINE TEXTURE AND ORIENTATION

DAVID N. LAMBETH, WEI YANG, HENG GONG, DAVID E. LAUGHLIN, BIN LU, LI-LIEN LEE, JIE ZOU, and PETER S. HARLLEE

ECE and MSE Departments, Data Storage Systems Center, Carnegie Mellon University, Pittsburgh, PA 15213, lambeth@ece.cmu.edu

ABSTRACT

Currently, and during the past few years, areal recording densities have doubled approximately every 18 months. This has resulted in the recent introduction of products exceeding 4 Gigabit/in^2. Furthermore, there have been recent laboratory demonstrations of over 11 Gigabit/in^2. This rapid technological pace is forcing media producers to seek new methods to control the media magnetic properties. It has recently been pointed out that due to the required signal to noise ratio and the resulting continued reduction in grain size the industry will soon be faced with the onset of thermal instabilities to data retention. Since the medium properties could limit the areal density of most recording systems, a systematic design approach toward media invention is necessary. Very small magnetic grains of near crystalline perfection will be required in order to achieve the coercivities and noise requirements for the next doubling of areal density (25 Gigabit/in^2). Following this not only will crystalline perfection be required, but extremely uniformly sized and singly oriented grains will be required to approach 50 Gigabit/in^2. Over the past few years we have taken an approach of controlled growth of the magnetic microstructure of thin film media to accomplish this. Here we provide an overview of the guiding media design philosophy and discuss materials issues, multi-layered thin film material structures and processing techniques which are used to control the microstructure and magnetic properties of Co-alloy films. Efforts toward epitaxial growth of multiple thin film layers on single crystalline Si is discussed as a method of achieving perfect crystallites of various highly oriented thin films. In the case of non-cubic materials, such as magnetic hcp Co-alloys, these films have well defined axial directions determined by the substrate and multiple thin film epitaxial relationships.

INTRODUCTION AND MOTIVATION

The continuing market force driven need for improved hard disk data storage is evident by the sudden change in the compound annual growth rate of areal densities in the early 1990's from 20 to 60%. This rate is still approximately 60% and there are no signs of abatement. The single largest, obvious, technical factor making this possible was the change from particulate disk media to sputtered thin film media. This smoother media surface has allowed continual decreases in the head to disk fly height to the present 50 nm and less. During the mid 1990's, even as trackwidths were shrinking, head disk interface technologies such as robust 15 nm thin CNx or CHx overcoats, laser produced mechanical texture and quasi-contact slider operation enabled reduced head to disk spacing and enabled the ubiquitous inductive head transducer to continue to perform adequately. With the advance to the more sensitive, low noise, magnetoresistive, and recently the spin valve device, record-playback transducer technology is returning media noise to being the limiting factor to further increases in areal density. While various laboratory demonstrations, such as IBM's recent announcement of 11 Gigabit/in^2 recording, have been used to herald the next generation of recording densities a more accurate measure of progress is provided by monitoring

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actual product performance. As of early 1998, the highest areal density of a commercial product fell very close to the 60% growth curve at 4.1 Giga bit/in^2. Concurrent with the progression of areal density has been a remarkable increase in data transfer rate and disk rotational frequency to decrease access times. This bandwidth increase, along with the introduction of the spin valve transducer, and the higher areal densities conspire to require even lower noise media.

These recent commercial improvements in longitudinal recording could not have been possible without the significant improvements already made in the recording media magnetic and microstructural properties. Ultimately, the achievable areal recording density should be determined by the media signal to noise ratio, which if the recording system is designed properly, is largely determined by the media thin film microstructure. For ideal media where the magnetic particles are totally non-interacting one might roughly estimate the media signal power to noise power ratio, SNR_{op}, (zero to peak signal to rms noise ratio) to be proportional to the average number of particles, N, sensed by the recording transducer. This assumption is based upon the concept that the large number of particles in a given volume is described by a Gaussian probability distribution and to determine the noise one merely estimates the variance to the average number of particles sensed. The signal power then goes as N^2 while the noise power goes as N. Hence, for a differentiating transducer system, 1000 particles would imply a SNR_{op} of 27db, a value that is sometimes viewed as necessary for an acceptable error rate. Therefore, assuming sufficient transducer sensitivity, in order to maintain an adequate SNR as the areal density is doubled, N would remain constant and, the number of particles per unit area would need to double. For thin film media, in which the particle extends through the thickness of the thin film, the particle or grain surface diameter would have to decrease by the square root of two.

Charap and Lu [1] recently modeled the limits of areal recording density based upon the concept that if a magnetic particle's anisotropy energy density-volume product is made too small it will spontaneously reverse due to thermal fluctuations. Their estimate of the limit of this product divided by the Boltzman energy is $K_uV/k_BT > 60$. At the same time since the media coercivity is proportional to the anisotropy field, $H_c \alpha H_k = 2M_s/K_u$ the anisotropy energy density cannot be increased to a point where Hc would be greater than the maximum available transducer record fields. These field levels are determined as a fraction of the Ms of available head transducer materials. Based upon these boundary conditions: a maximum K_u to allow recording, a minimum thermally stable grain volume and a fixed number of grains to provide the required SNR, they estimated that a limit to long term data stability may occur around 40 Gigabit/in^2. We would like to suggest, however, that the statistical arguments concerning the number of particles required may be somewhat flawed. Unlike particulate tape media, thin film media is essentially volume filling. That is, there are no intentional voids in the media and the total magnetic moment sensed by the transducer would be essentially constant if the magnetic easy axes of all of the grains were in the same direction. In traditional rotating thin film hard disk media the easy axis is designed to be random in the plane of the disk to avoid modulation of the signal as the disk turns. Hence, at any position on the disk a large fraction of the grains have their easy axes either perpendicular or considerably off axis from the transducer recording direction. In the extreme case of totally non-interacting particles these grains contribute little or no output flux to the transducer and appear as magnetic voids. Hence, even if all particles were the same size and regularly spaced, mis-orientation would still provide a mechanism for fluctuation in the signal. Clearly these mis-orientations along with the distribution of grain sizes limits the SNR. By narrowing the size distribution or, better still, restricting it to a singly uniform size the SNR would improve. However, by orienting all the easy axes to a single direction, comparable to the transducers' sensitive direction, the noise would become dependent only upon the size distribution and would be further reduced. A smaller number of grains would be required to achieve the same

SNR and recording densities could be extended beyond the current suggested thermal stability limit. Since grain size distributions are typically skewed and the smaller particles are thermally unstable they should be eliminated. By eliminating these and the variance in particle number due to random orientation distribution it would seem to be reasonable to expect that the noise could be reduced by at least a square root of two and perhaps considerably more. A factor of two reduction would lift the predicted areal recording density limit to well over 50 Gigabit/in^2.

The above arguments assume non-interacting particles and it would seem to be naive to ignore magnetostatic or intergranular exchange interactions in a discussion of noise. However, it is the goal in the media design to eliminate the intergranular exchange by isolating each grain. Hence, noise induced by magnetostatic interactions can be limited if the media and recording system are designed correctly, as discussed next.

MAGNETIC RECORDING PHYSICS ISSUES

While the playback signal is extremely dependent upon the head to medium spacing and transducer resolution (gap length), it is also proportional to the magnetic medium film thickness and the spatial gradient of the magnetization along the recording track. Hence, one would be inclined to argue that the larger the remnant-film thickness product ($M_r\delta$) the better. This is not the case however, as the recorded transition length increases with this quantity and magneto-resistive playback transducers are easily driven into a non-linear regime of operation if the sensed field is too large. Hence, there is a maximum, and optimal $M_r\delta$, for the media determined by the head sensitivity function and the head to media spacing. For currently commercial high performance media $M_r\delta$ has been reduced to less than 0.5 milli-emu/cm^2. This is beneficial to media design for two reasons: One is that, for exchange coupled media or for a system with a poor record head field gradient, the apparent media transition noise can be directly correlated to the demagnetization fields at the transition. The second reason is that the down-track (linear) recording density is limited by the finite flux reversal length. This length is measured by the transition parameter, a_x, which is determined either via transition demagnetization forces or by the finite record head field gradient. The media noise is always lower if the coercivity is greater than the demagnetization field (Hc > Hd). In the ideal limit where the head is in contact with an zero thickness media and has a zero gap length the recorded magnetization profile would be a step function provided that Hc > Hd and that the media were homogenous. For the more realistic media microstructure the transition length would be determined by the characteristics of the set of grains that lie at the immediate location where the ideal transtion would have been. The transition length is then nominally the average grain size and the transition location variance (noise) is determined by the grain location, size and orientation distributions. In other words, even if media microstructure were made ideal, it would still be easy to incorrectly design a recording system to induce apparent media noise.

As an example for disussion consider Figure 1 which shows the media noise power spectra of several differing media determined by the noise power spectral integration technique. These datum were obtained using a Read Rite Tripad head with a .22 micron gap and flying just above medium contact at approximately 25 nm (at 7.1 m/sec.). In each case it is noted that the noise power initially increases linearly with density as the noise power is dominated by the flux transition location jitter and increases linearly with recording frequency as transition density increases. The fact that there is noise even at zero frequency indicates the discrete non-zero size of the randomly oriented magnetic switching units or grains. The initial slopes of these curves are determined by the grain size, intergranular exchange coupling, the maximum demagnetization field (determined by $M_r\delta$), and the finite head field gradient. At higher recording densities it is common that the noise increases supra-linearly as the individual flux transitions interact during the recording process and begin to interpercolate. The onset of this supra-linearly noise behavior typically occurs as the flux transition density approaches the transition length, a_x. From a practical standpoint, but to some extent dependent upon the signal processing encoding technique, this noise limits the data transition spacing to approximately πa_x. At extremely high flux transition densities the noise again decreases as the media appears to be AC erased. A media with highly exchange coupled grains will show little DC frequency noise as the magnetization of all grains tend to align, while when AC erased the percolation effect is large. This DC result is a clear example that thin film media is volume filling and that this noise is not simply controlled by particle counting statistics. For a media with little exchange coupling the DC noise will correlate to the number density and orientation of the grains. When the media is designed with very small grain sizes and such that the Hd < Hc the head field gradient during the record process determines a_x and not the demagnetization fields associated with the media. The lowest noise curves of Figure 1 indicate that these media have sufficiently small grains (low slope) and a small enough $4\pi M_s$ /Hc that the supra-linear increase in noise does not occur until well after the maximum measured kfci. Hence, the grain size (isolated magnetic switching unit size) in combination with the random in-plane orientation determines the slope of this curve, as well as, the DC erased noise power shown at low frequencies. Whereas, media with large grains will show a large low flux density noise slope and if the Hd > Hc the supra-linear noise behavior will set on early. The higher noise curves represent this type of media. The lowest noise media of this set is comparable to today's best commercial media. If a medium noise is not head field gradient limited, but limited

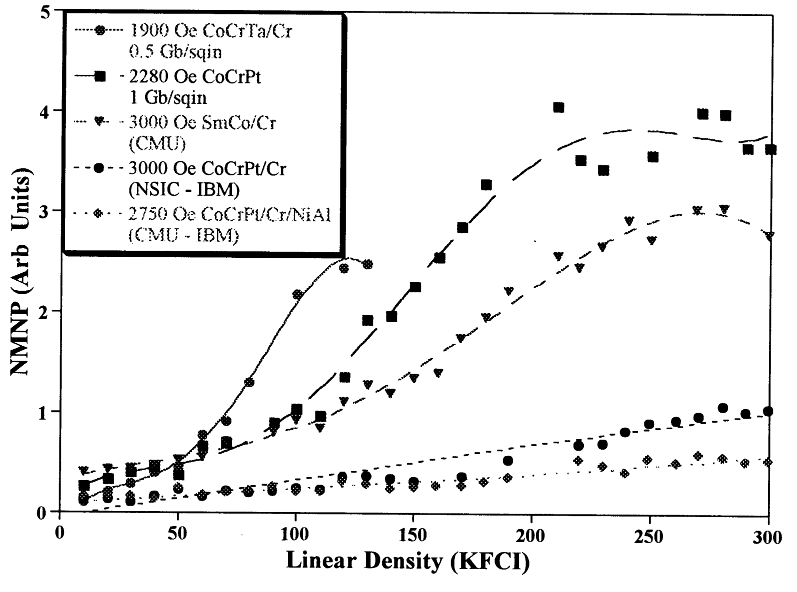


Figure 1. Normalized media noise vs. flux reversal density for media with various degrees of exchange coupling, Mrδ, and head field gradient induced noise.

by the statistical nature of the medium grains, then the required average grain size, assuming random size and orientation distributions, for 10 Gigabit/in^2 technology is estimated to be about 15-20 nm.

MEDIA CONSTRUCTION AND THE ROLE OF CRYSTALLINE TEXTURE

To attain the highest recording densities the coercivity should be maximized. In order to accomplish this the media designer could choose a magnetic alloy to achieve a higher anisotropy constant, lower the $4\pi M_S$ which also decreases the flux transition demagnetization effects, or gain better control of the microstructure to achieve the maximum potential anisotropy energy density of a given alloy. While studies have been performed on other alloys, in modern hard disk media hcp Co alloys are almost exclusively chosen due to their corrosion resistance and high anisotropy constants. Second and third elements such as Ni, Cr, Ta, Nb, B or Pt are chosen to promote diffusion of non-magnetic elements to the grain boundaries during film processing. In current products Cr and Ta are widely used and have proven to be especially useful at providing magnetic grain to grain isolation, while Pt appears to increase the anisotropy constant, but is not as effective at providing isolation. Likewise, these non-ferromagnetic elements usually lower the magnetization by dilution. The Co alloy is typically deposited onto a thin film underlayer structure which induces both an hcp Co phase and orients the crystalline c-axis by epitaxial growth. Perfect, defect free and isolated, hcp crystalline grains of appropriate size insure that domain walls do not nucleate at grain boundaries, crystalline flaws or stacking faults to lower the coercivity, while the orientation of the c-axis determines the maximum achievable coherent rotation coercivity. Via modeling, Yang [2] has predicted the hysteresis loop dependence upon the orientation of the c-axis with respect to the film plane. For ideal Stoner-Wohlfarth particles with easy axes (c-axes) parallel to the applied record field the coercivity would be equal to the anisotropy field, $H_k = 2K_u/M_S$, while for a random ensemble of particles with c-axes in the film plane the predicted coercivity is reduced to 0.51 of H_K. However, if the c-axes are randomly oriented in all three dimensions the grains with axes out of plane, (or even only somewhat dispersed about the plane) have their magnetization forced back into the plane via demagnetization fields and these additional fields further reduce the coercivity. For the rotating longitudinal recording hard disk media format the singly directed ensemble is currently impractical and so the random two-dimensional structure is the most desirable. Hence, the choice of the underlayer texture upon which to perform epitaxial growth is critical in determining the Co alloy c-axis orientation. In addition these underlayer structures are critical in determining Co alloy crystalline quality and the grain size. Historically [3], and even though a number of other elements have been investigated, bcc Cr and Cr alloys have been used almost exclusively for this purpose.

Most current hard disk thin film magnetic media are constructed upon a highly polished NiP electrolessly plated AlMg, glass, or glass ceramic substrate by sputter deposition of a sequence of metallic layers. The exact structure depends upon the substrate, but usually consists of sequential depositions of a non-magnetic seed layer and underlayer followed by a magnetic Co-alloy, followed by a ceramic-like protective coating (principally carbon, CH_x or CN_x), and finally a very thin lubricant. The seed layers that have been used include both oxides and metals depending upon the substrate and the manufacturer. The purpose of the seed and the underlayers are to buffer the substrate surface and to initiate the crystalline growth and texture of the very thin magnetic layer. Hence, their composition, interaction with the substrate, and the processing conditions are important in determining the microstructural interface to the magnetic layer. In addition, the objective of the seed and underlayer is to establish a controlled grain size for the growth of other epitaxial layers.

If the atoms of a metallic thin film have sufficient surface mobility during the deposition process they arrange into a minimum surface energy configuration. For most simple lattice structures this implies a close packed atomic surface configuration. For fcc or hcp lattices the thin film surface will have a (111) or (0002) texture, respectively, while for the bcc lattices a (110) texture results. Hence, for bcc Cr alloys the (110) texture is commonly observed. As an example of this process consider Figure 2 which shows a series of x-ray diffraction θ-2θ scans of a CoCrPt film sputter deposited upon very thin, but varying thicknesses of sputter deposited Cr on a glass substrate. The glass substrate represents a high energy oxidized surface upon which arriving Co atoms would have limited mobility. By first placing a very thin Cr layer on this oxidized surface the mobility of the Co atoms can be significantly increased. However, below 5 nm the Cr is so thin that little or no crystalline texture has evolved and so it could, perhaps, be thought of as amorphous or very conducive to strain relaxation. Nevertheless, its presence alters the interfacial energy between the glass substrate and the Co alloy sufficiently to allow the thicker Co alloy film to seek its lowest energy close packed arrangement. Hence, while the (0002) texture is initially

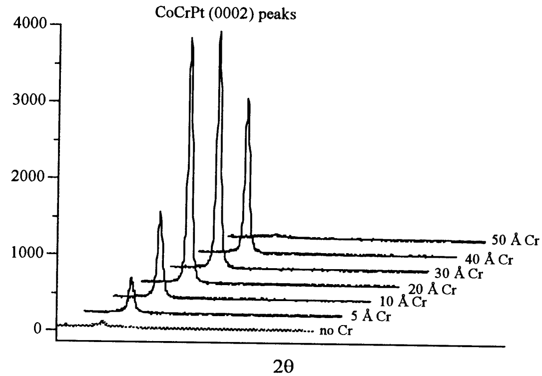


Figure 2. Close packed (0002) texture evolution of 40 nm thick CoCrPt grown on very thin Cr.

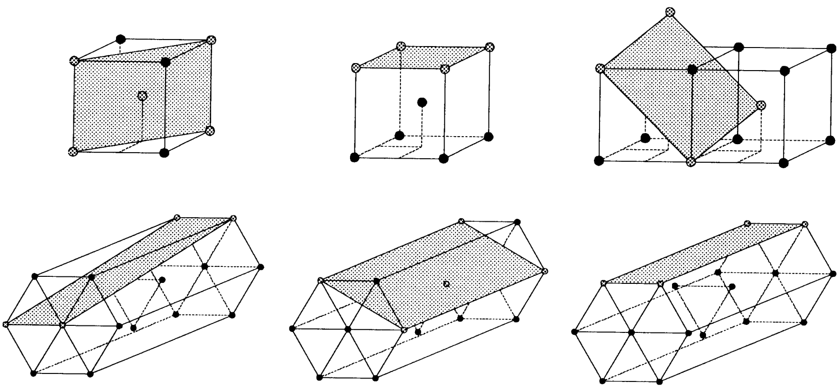


Figure 3. Epitaxial orientational relationships between bcc Cr and hcp Co.

not present when little or no Cr is deposited, it appears quite strong as the Cr thickness becomes sufficient to allow the Co to wet the substrate, then again disappears as the Cr becomes thick enough to develop and maintain a texture upon which the Co alloy is forced to epitaxially grow in different directions. The fact that the Cr has not developed a strong texture of its own at 5 nm thickness results in the Co texture being random. A similar effect, of this natural Co texture evolution, would occur if the underlayer allowed wetting of the Co to the surface or did not present a crystalline structure suitable for any epitaxial growth. Ta films, which appear amorphous when very thin, are known to perform a similar function for (111) fcc Ni alloys and are typically used as initializing seed layers for spin valve transducers. Other crystalline metals will perform this task if the lattice spacing or crystalline symmetry is largely different from that of the following layer. Likewise, if the atomic mobility is limited by substrate to film interfacial energy or by competition between the deposition rate and the atomic surface relaxation, or by impinging atomic kinetic energy, then other textures may appear. Consequentially, if the film nucleation process produces island like growth with high aspect ratios then the sides of the islands can represent a large fraction of a nucleating grain surface area and the lowest surface energy {110} planes of a bcc will not be parallel to the substrate surface, but to the island sides. This results in a (002) texture in addition to the common (110) texture. Historically the most commonly sought Cr texture has been the (002). To a limited extent this texture can be induced by deposition at elevated temperatures and at high deposition rates which induce a high aspect ratio island growth [4]. On occasion, especially for very thick films, as a powder diffraction pattern might begin to appear, the (112) texture would even be observed. Hence, the epitaxial growth relationships between Co alloys and the various Cr textures have been discussed in depth [5 and references therein]. The more relevant texture and orientation relationships are summarized as:

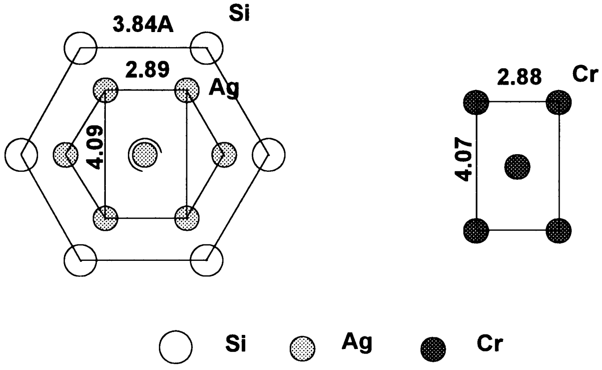
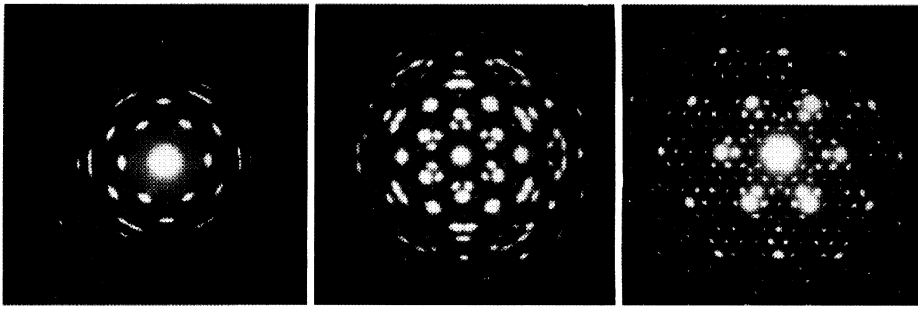


Figure 4. Orientational relationships for Cr(110)/Ag(111)/Si(111).

Bi-crystal:	$\mathrm{Co}(11\underline{2}0)[0001] \parallel \mathrm{Cr}(002)[110]$ or $\mathrm{Co}(11\underline{2}0)[0001] \parallel \mathrm{Cr}(002)[1\underline{1}0]$
Quad-crystal:	$\mathrm{Co}(10\underline{1}1)[\underline{1}2\underline{1}0] \parallel \mathrm{Cr}(110)[1\underline{1}0]$ or $\mathrm{Co}(10\underline{1}1)[\underline{1}2\underline{1}0] \parallel \mathrm{Cr}(110)[\underline{1}10]$
Uni-crystal:	$\mathrm{Co}(10\underline{1}0)[0001] \parallel \mathrm{Cr}(112)[1\underline{1}0]$

Figure 3 illustrates these three Co textures. While the Cr (002) and Cr (112) textures induce the Co c-axis into the film plane the most easily formed Cr (110) texture results in the c-axis being inclined at ±28 degrees with respect to the surface. Hence, a lower coercivity would be anticipated from the Co grown on the Cr (110) texture as the c-axis is not parallel to the recording plane. Also we see that there are multiple directions that the Co c-axis can be placed upon the Cr (002) and the Cr (110) textures. Hence, upon a single (002) textured Cr grain two possible c-axis orientations can grow (bi-crystal) while upon a single (110) textured Cr grain four



Cr(110) **Cr(110)/Ag(111)** **Ag(111)/Si(111)**

Figure 5. TEM diffraction patterns for Cr(110) and Ag(111) epitaxially grown on Si(111).

possible c-axis orientations of Cr can grow (quad-crystal). These bi-crystals and quad-crystals can never have both c-axes parallel to the applied field simultaneously. Hence, by the Stoner-Wolhfarth model one would anticipate the coercivity to be compromised. Likewise, when two or more Co variants do appear on a single Cr grain it is less likely that they will be isolated from one another by grain boundary diffusion as when compared to two Co grains located on two separate Cr grains. Hence, the boundary between the two variants can be thought of as a crystalline defect at which the magnetic spin orientation must be twisted and this provides for wall nucleation or an incoherent spin rotation center to initiate the switching process. This also compromises the maximum achievable particle anisotropy and hence, coercivity. Even worse, the four possible directions of the Co(10$\bar{1}$1)//Cr(110) texture relation places the Co c-axes ±28 degrees from the film plane and this in combination with the perpendicular demagnetization field always lowers the coercivity to an in-plane field. On the other hand, in the absence of severe Co alloy compositional flaws, the uni-crystal Co(10$\bar{1}$0)//Cr(112) texture relationship only allows a single orientation upon a Cr grain and conceivably could result in a higher coercivity if aligned to the recording field. In addition, the surface atomic spacings (a/sqrt(3) = 2.50; sqrt(2) x a = 4.07A) of this Cr texture closely matches both the c (~4.07A) and a (~2.50A) axes lattice spacing of the Co alloy simultaneously while the other Cr textures only match well in one of the two directions. Since only an uni-crystal can grow on an individual underlayer grain the anisotropy energy is not compromised by multiple growth variants. Unfortunately, this Cr texture is seldom seen, as processing at low temperatures or with an applied substrate bias results in Cr (110) texture while high temperatures and deposition rates can partially induce the Cr (002) texture. It is believed that the (112) texture of the bcc derivative, NiAl, provides a high coercivity template while the strong intermetallic Ni-Al bonding provids a small uniform grain size template [6,7].

In addition, it should be mentioned that extended Co alloy stacking faults, caused by compositional inhomogeneity or epitaxial lattice mis-match, could appear as small regions of the Co fcc phase. This cubic phase has a considerably lower anisotropy than the hcp phase and since it is in intimate exchange contact with the remaining hcp portion of the crystallite it may locally reduce the anisotropy energy and, hence, the coercivity. Processing at an elevated temperature and possibly with substrate bias during the Co deposition helps to provide the atomic mobility to minimize this crystalline disorder. The first criterion for the selection of Cr as an underlayer for Co was the close atomic lattice spacing match. Hence, these stacking flaws are exaggerated, or epitaxial growth does not even result, if the Co-alloy additives create too great a lattice mis-match. Due to its large atomic size Pt solutes significantly increase the Co lattice constant and to

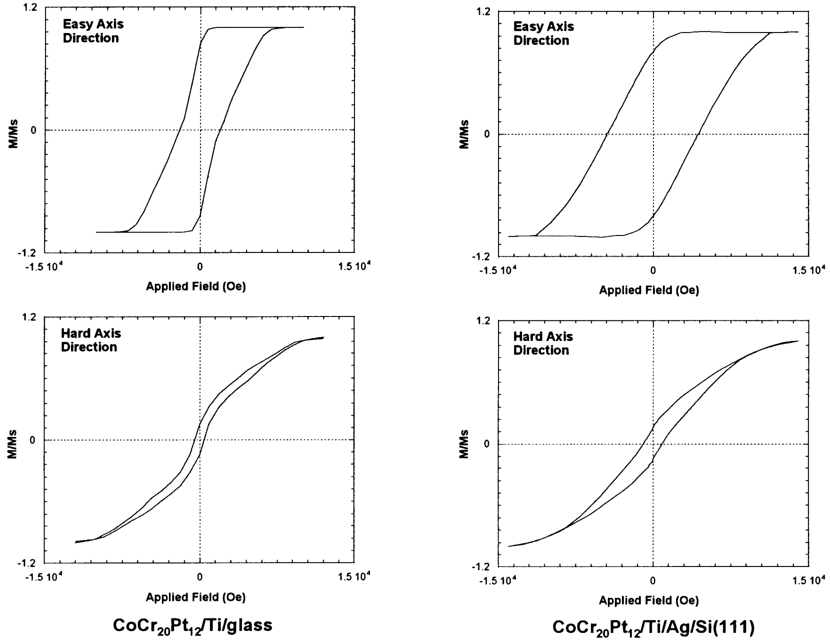


Figure 6. Comparison of easy and hard axes hysteresis loops for CoCrPt (0002) films.

correct for this V, Ti, and Mo have been alloyed into Cr to expand its lattice constant appropriately.

HIGHLY ORIENTED MAGNETIC THIN FILMS

From the previous discussions, we see that it is desirable to grow Co-alloy and underlayer crystalline grains of considerable perfection and appropriate texture to avoid compromising the crystalline anisotropy. Likewise, to maximize the coercivity and to minimize the media noise it is desirable to control the easy axes orientations. By utilizing single crystal Si as a substrate we have developed a model system to approach these goals. Silicon is reasonably inexpensive and could conceivably be used as a media substrate. By first removing the SiO_2 surface layers a metallic epitaxial layer growth can be obtained. In particular we have found that fcc metals such as Ag, Au, Cu, and Al can grow on various Si surface orientations with a very high degree of epitaxy. Furthermore, these fcc metallic quasi-single crystal thin films can then easily be used to epitaxially grow other quasi-single crystal films of similar or differing crystalline structure. By proper choice of lattice constants various textures can be achieved with limited induced lattice strain. Multiple layers allow a transition from what would be thought of as an impossible lattice strain situation to one of little strain, a high degree of texture and orientation results as determined by the substrate [8]. As examples consider a few of the texture relationships we have obtained:

Bi-crystal: Co(11$\underline{2}$0)/Cr(100)/Ag(100)/Si(100)

Quad-crystal: Co(10$\underline{1}$1)/Cr(110)/Ag(111)/Si(111)

Perpendicular: Co(0002)/Ti(0001)/Ag(111)/Si(111)

Soft cubic: NiFe(111)/Cu(111)/Ag(111)/Si(111)

Uni-crystal: Co(10$\underline{1}$0)/Cr(112)/Ag(110)/Si(110).

In each of these the Ag to Si lattice match occurs at a unit cell ratio of 4 to 3, respectively. The very long range order of the single crystal Si is believed to promote the order of the metal contacting layer. The following metal layers' crystalline axes may then either align to the principle Si axes or be rotated through fixed angles depending upon the lattice parameter relationships. In the case of Co growth on the quasi-single crystal Cr thin films the orientational relationships listed in the earlier discussion of texture apply. The excellent epitaxial relationships between the single crystal substrate and the bi-crystal Co(11$\underline{2}$0) were described earlier[8], however, it was not clear from that work that the extensive array of other epitaxial relationships would exist.

Consider Figure 4 showing one of the three orientational relationships of bcc Cr(110) on fcc Ag(111) used to promote the quad-crystal structure. It is worth noting that these are the close packed lattice planes and since the atomic spacings of the Cr and the Ag match reasonably well the high energy Cr(111) would not be anticipated to grow. Since there are three possible orientations for the Cr(110) on the Ag(111) surface and since there are four possible orientations (quad-crystal) for the Co(10$\underline{1}$1) growth on Cr(110) there are actually twelve possible Co orientations on the Si substrate. This represents enough orientations that it is tempting to advocate that a quad-crystal rotating disk made from a Si(111) substrate would have little rotational signal modulation. However, this would be far from the uni-orientational structure

NiFe 50nm/Cu 100nm/Ag 100nm/Si(111)

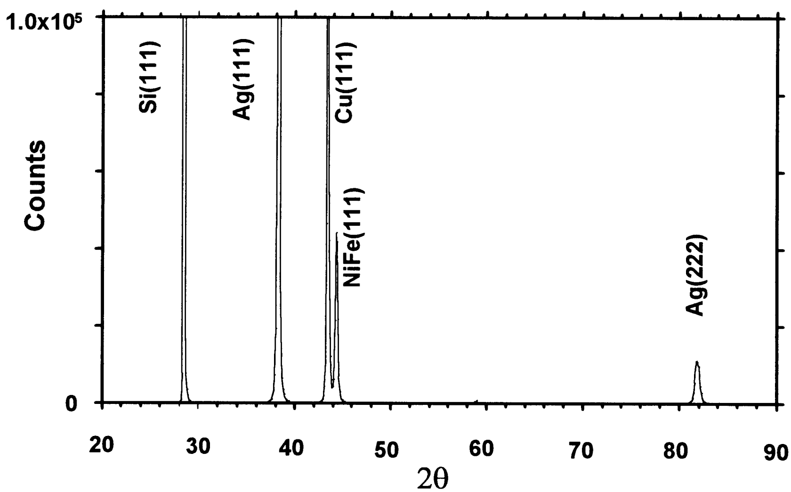


Figure 7. X-ray diffraction scans of NiFe(111)/Cu(111)/Ag(111)/Si(111).

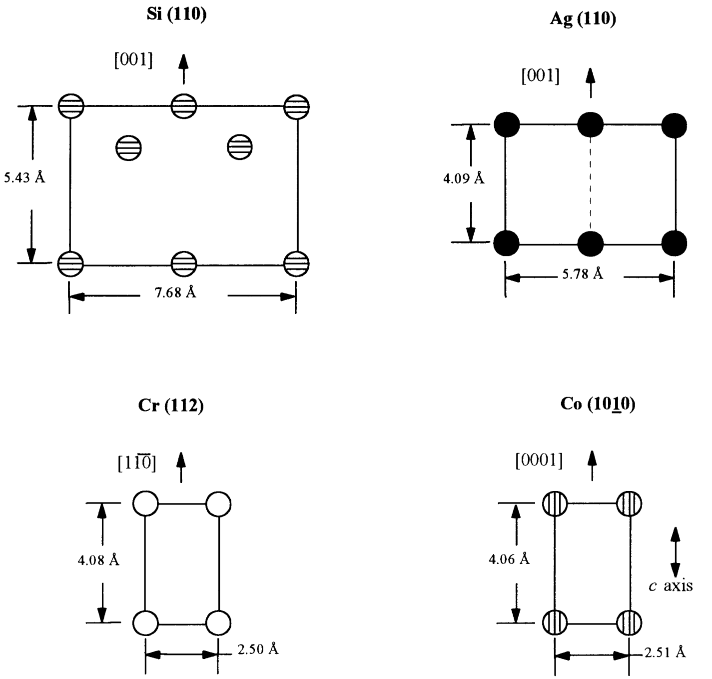


Figure 8. Orientation relationships for the uni-crystal Co(10$\underline{1}$0)/Cr(112)/Ag(110)/Si(110).

advocated earlier. Figure 5 shows TEM electron diffraction patterns to demonstrate the alignment between the Si substrate and the metal underlayers for the quad-crystal structure.

In conventional perpendicular recording media Ti is commonly used as an underlayer to induce the Co-alloy (0002) texture. However, the Co-alloy(0002) texture that develops on Ti usually requires that the Co-alloy be grown rather thick. Since Ti(0001) (a = 2.95A) is well lattice matched to Ag (a/sqrt(2) = 2.89A) highly textured Ti(0001) films were grown on quasi-crystal Ag(111)/Si(111) templates. Unfortunately, the Ti(0001) does not provide a good lattice match for Co(0002) (a = 2.50A). Nevertheless as an example of the effect of providing a long

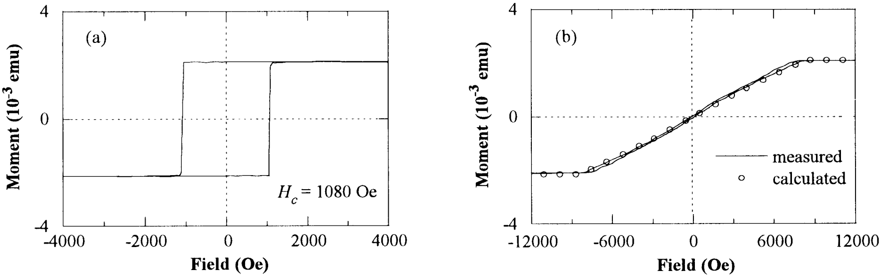


Figure 9. Easy and hard axes hystersis loops for uni-crystal Co-alloy films.

range ordered underlayer Co/Ti was grown on Ag(111)/Si(111) and for comparison to a conventionally prepared perpendicular media it was simultaneously deposited on glass. Hystersis loops for both the easy and hard axes of both samples are shown in Figure 6. Clearly the magnetic characteristics are much improved when the Si substrate is used.

For perpendicular media a soft magnetic underlayer is commonly advocated. Figure 7 shows the $\theta-2\theta$ x-ray diffraction scans for an epitaxial (111) textured soft NiFe layer. The diffraction peak heights are extremely large, especially for the Cu and Ag, which are buried under the NiFe, when compared to films which would be grown without the single crystal Si substrate. Peak heights of thousands of counts are common for most of the films grown on Si, whereas a typical Cr(110) film grown on glass and measured in the same manner would have peak heights of only tens of counts. Rocking curve measurements of each of the Ag(111), Cu(111) and NiFe(111) film layers indicate a very small dispersion ($\Delta\theta_{50} \leq 0.7°$), while preliminary TEM data indicates epitaxial orientation of each layer. The hard axis coercivity of approximately an oersted and is believed to be due to the influence the earth's field during deposition.

For the often sought after uni-crystal structure Cr(112) develops nicely on the Ag(110)/Si(110) surface. TEM diffraction patterns show a clear epitaxial relationship as detailed in Figure 8. By calculation, one would predict that Cr(110) would grow on Ag(110) because a good lattice match is found in two directions. However, while Cr(112) only matches in one direction, it is suspected that extra energy due to the lack of a bond for the body-centered atom dominates and forces the (112) texture. The resulting Co films are highly oriented and represent a wonderful opportunity for the study of anisotropy energy. Figure 9 shows the nearly ideal easy and hard axes loops where the calculated data points were derived from measured torque curves.

CONCLUSIONS

In this paper we have presented a brief summary of the magnetic recording physics issues that motivate better crystalline quality, texture, and orientation of thin film hard disk media of the future. Arguments concerning the postponement of reaching thermal stability limits to magnetic data retention were given to justify the need for achieving a singly oriented media. We then gave a summary of new methods of achieving highly oriented magnetic materials via epitaxial growth on single crystal silicon.

The results presented represent knowledge accumulated from many years of effort by students, staff, and faculty involved with media development at Carnegie Mellon University, as well as discussions with many other friends and colleagues. Much of the work described was supported in part by the Department of Energy (DE-FG02-90ER45423), the National Science Foundation (ECD 89-07068), or via grants from IBM, Intevac, and Seagate.

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