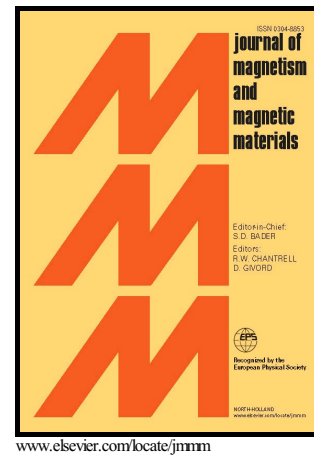


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Preparation and Characterization of Bottom Ferromagnetic Electrode for Graphene Based Magnetic Junction

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Abstract

Magnetic multilayer stacks incorporating several layers of graphene have been predicted to produce very high magnetoresistance and high conductivity, a combination of properties that would be useful in magnetic sensors and future spin-based data storage and processing technologies such as MRAM. To realize the theoretically modelled heterostructures and probe their properties, a clean, high-quality graphene-ferromagnet interface, such as one that results from CVD of graphene directly on ferromagnetic films, is required. However, past works using Ni and Co films for CVD of graphene employ the ferromagnetic film as a sacrificial layer to be dissolved after graphene growth and ignore changes to its morphology and magnetic properties. Here we investigated the effect of graphene CVD growth conditions on the properties of Co, Ni, $\text{Co}_{90}\text{Fe}_{10}$ and $\text{Ni}_{80}\text{Fe}_{20}$ ferromagnetic films. The magnetic films were grown by dc magnetron sputtering with different growth conditions onto c- Al_2O_3 , Si/AlN and MgO substrates. The crystalline orientation, surface morphology/roughness and magnetic properties of the films were measured using X-ray diffraction, atomic force microscopy and vibrating sample magnetometry, respectively. Cobalt films grown at 500° C were found to be hcp and heteroepitaxial on c- Al_2O_3 . CoFe, Ni, and NiFe films on c- Al_2O_3 were found to be fcc and to be (111) textured but with grains having in-plane rotation differing by 60°. The CoFe and NiFe films on c- Al_2O_3 retained their small coercivity and high remanence while the pure Co and Ni films exhibited much smaller remanence after graphene growth, making them unsuitable for magnetic memory technologies. Films on Si/AlN were found to have the same rotational domains as those on sapphire c- Al_2O_3 . The NiFe films on (111) MgO were found to be mostly single domain.

Keywords:

Magnetic junction, graphene, magnetic electrode, chemical vapour deposition.

Introduction

High-quality graphene-ferromagnet vertical heterostructures have been predicted to possess a high conductance and high magnetoresistance, a combination of properties useful in future spin-based

technologies like magnetic random access memory (MRAM), spin-based logic and magnetic sensors. The prediction is based on a comparison of the spin-polarized band structures of Ni, Co and graphene (Gr) and the close lattice match of the three close-packed surfaces: Ni and Co possess a majority-spin band gap at the K point in the first Brillouin zone while graphene only has states near the K point at low energies. As a result, graphene acts as a tunnel barrier for majority-spin electrons while minority spin electrons may pass through the trilayer with little resistance, giving rise to extraordinary magnetoresistance. The predicted spin polarization is 25% for a perfectly ordered single crystal Ni (111)/Gr/Ni(111) junction with a single graphene layer and it exceeds 2400% for greater than four graphene layers. Spin-polarized tunnelling through a single layer of graphene has been observed experimentally by several authors. Cobas et al. [3] reported a tunnel magnetoresistance (TMR) effect of 2.5% for a $\text{Ni}_{80}\text{Fe}_{20}/\text{Gr}/\text{Co}$ magnetic tunnel junction (MTJ) structures using a single layer graphene transferred from a Cu foil after chemical vapour deposition (CVD) growth. Similarly, low temperature magnetoresistances of 1.5-3.4% were reported for a NiFe (Py)/Gr/Co magnetic tunnel junction (MTJ) by depositing the two magnetic electrodes onto a free standing single layer of graphene [4]. Separate experiments have confirmed the minority spin-filtering nature of the graphene-nickel interface [5].

However, none of these experiments have realized the clean, high-quality crystalline graphene-ferromagnet interface envisioned in the theory or achieved the extraordinary properties predicted. To achieve better performance a clean, high-quality, crystalline realization of a ferromagnet-graphene interface is required. Although many works have achieved and studied chemical vapour deposition of graphene on ferromagnetic surfaces, the studies treated the ferromagnetic film or foil as a sacrificial catalyst and have not studied the physical and magnetic properties of the film that result from the heat treatment conditions present in CVD graphene growth. To realize an accurate reproduction of the theoretically treated FM-Gr-FM trilayer via CVD, careful preparation of the ferromagnetic film is required and its magnetic properties must be understood before and after the heat treatment process used to grow graphene.

Preparation of epitaxial graphene on metal surfaces is typically by C-deposition, by segregation of C from the bulk of the metal to its surface, or by chemical vapor deposition of hydrocarbons [6]. Single layer graphene (SLG) or few layer graphene (FLG) can be grown on Ni and Co by CVD at high temperature, using various hydrocarbon precursors. Unlike Cu or Ru where the graphene growth is a surface catalytic reaction, the growth of graphene on Ni and Co is by precipitation upon cooling, due to the large C solubility at the CVD temperature [7]. A recent review article stated that single layer or bi-layer graphene are often rotated relative to the Ni (111) lattice [8]. Synthesizing uniform defect-free multilayer graphene lattice matched to the substrate metals is even more challenging, and the growth mechanisms are less understood. It requires careful control of metal thickness, temperature, annealing time, cooling rate, the metal composition and microstructure, and also the type of hydrocarbon species and CVD pressure [9]. The amount of carbon segregation during cooling is limited by the total amount of carbon dissolved in the metal film, which is determined by the film thickness and the growth temperature. In order to form 3 to 5 layers of graphene, a typical thickness of ~300 nm of Ni film is used [7,10].

Co and Ni are ideal candidates as the bottom magnetic electrode for graphene based MTJs as they only have a small lattice mismatch with graphene. Fcc Ni (111) can grow epitaxially onto c- Al_2O_3 but usually it has a two-domain structure with 60° of in-plane rotational disorder [11]. This is due to the two types of fcc stacking sequence, ABC and ACB generating two equivalent domains. The locally enhanced carbon segregation at the domain boundaries is the origin of inhomogeneous graphene growth on (111) Ni film [7]. Better growth control and better uniformity of graphene are expected if grain boundaries are eliminated. Single crystalline Ni films grown on c- Al_2O_3 have been reported by

careful control of deposition parameters [12]. On the other hand, hcp Co film on c-Al₂O₃ can be more easily grown as single domain [13]. The formation of complex multidomain structure of the Co film due to the fcc-hcp transition cooling below 422° C seems to result in the large inhomogeneities in graphene growth [7].

In this paper, we investigated growing an epitaxial bottom magnetic electrode on a variety of single crystal substrates and examined the properties relevant to fabrication of a magnetic FM-Gr-FM trilayer structure both before and after high-temperature heat treatment associated with graphene synthesis. We investigated both pure Ni and Co films as well as CoFe and NiFe alloys expected to have better magnetic switching characteristics. We deposited the films on lattice-matched substrates of c-Al₂O₃ as well as (111) MgO and (111) Si/(0001) AlN substrates expected to improve crystallinity and ease integration with current semiconductor manufacturing processes.

Experimental Methods

All of the magnetic films were prepared using DC magnetron sputtering in an AJA sputtering chamber with base pressure less than 3×10^{-8} Torr. The sputtering guns were in a confocal geometry with $\sim 15^\circ$ of incidence angle and ~ 15 cm of distance between substrate and the gun. The substrate materials were ultrasonically cleaned 15 min each in acetone then alcohol. Single crystal c-Al₂O₃, (111) MgO and PVDNC coated (0001) AlN on (111) Si were purchased from commercial sources. Quartz, Si/SiO₂, or Si/Si₃N₄ substrates were sometimes used to grow reference samples. Substrates were heated to 750-800° C in vacuum prior to deposition. The heating was by halogen lamps from the backside of the sample stage. The thermometer was positioned at about 2 cm above the back of the substrate and so the real temperature of the substrate may deviate from the temperature reading by as much as 50 degrees.

The growth temperature was varied from 300 to 600° C, the Argon working gas pressure varied from 1 to 4 mTorr and the sputtering power from 100 to 200 W. RHEED was used to examine the surface ordering of the films after growth. X-ray measurements were done using a four circle PANalytical X'Pert MRD with Cu radiation. X-ray reflectivity was used to measure film thickness and film density while phase measurements, pole figures and rocking curves were performed to examine the structural and orientation information. In-plane and out-of-plane hysteresis loops were measured at room temperature using a MicroSense vibrating sample magnetometer at up to 20 kOe. The surface morphology and roughness were measured using an Atomic Force Microscopy.

The typical CVD growth process is to heat the substrates with the ferromagnetic film to 850-900° C in an Argon-Hydrogen ambient. High temperature is maintained for 10-30 min in Methane-Hydrogen atmosphere and the samples are cooled under Argon-Hydrogen atmosphere.

Results

I. Co films

Cobalt films of ~ 300 nm thickness were grown on Si/Si₃N₄, Si/SiO₂, c-Al₂O₃, and on top of Si/SiO₂/7nm Ta/10nm Ru. They were deposited at 200 W, 3 mTorr of Ar at 500° C and then annealed in 10 mTorr of Ar/8% H₂ at 500° C for 30 min before cooling. As expected, the films on Si/Si₃N₄ and Si/SiO₂ were polycrystalline, with a large rms roughness of ~ 14 nm. However, the Co films on Si with Ta/Ru as underlayers and on sapphire substrate both showed intense (0002) peak of hcp Co at 2 theta angle of 44.54° in the 2 theta omega scans (fig. 1), indicating a highly (0002) textured nature. The crystal orientation of these two films was further identified, using pole figure measurements, as randomly

oriented in the film plane for the former (fig. 2a) and single crystalline film for the latter, with the orientation relationship as (0001)Co//(0001)Al₂O₃ and [11 $\bar{2}$ 0]Co//[10 $\bar{1}$ 0]Al₂O₃ (fig. 2b). In comparison, the rms roughness for the film on Si/Ta/Ru is about 13 nm while it is only 0.5 nm on the sapphire. The smoothness of the Co film on sapphire is reflected in the small full width half maximum (FWHM) of 0.338° for its (0002) peak.

The single crystalline nature and the smoothness of the Co on sapphire film suggest this film should be a good candidate as the bottom magnetic electrode for the graphene MTJ. However, the measured in-plane M vs H loop showed a slanted slope (fig. 3a) with a large H_c of 139 Oe and M_r/M_s squareness of 0.21 (table 1), giving it poor switching characteristics for a spin valve structure. Nevertheless, the out of plane c-axis orientation renders it useful as a perpendicular magnetic anisotropy (PMA) material for small size devices.

II. CoFe film

We investigated the substitution of Co with Fe to mitigate this problem. At the composition of Co₉₀Fe₁₀ (CoFe), Fe stabilized the fcc phase which reduces the magnetocrystalline anisotropy of the alloy and makes it softer. Figure 3b shows the M vs H loop of a CoFe film grown on c-Al₂O₃ at 500° C with 100 W of power at Ar gas pressure of 3 mTorr. The H_c value was reduced from 139 Oe for the Co film to 4.6 Oe and the M_r/M_s squareness was increased from 0.21 to 0.97 (table 1). This property makes the film suitable to be used as the bottom magnetic electrode for magnetization in-plane graphene MTJ device.

The (200) pole figure (fig. 4a) at psi value of 54.7° showed this film is indeed face centered cubic. The six fold symmetry of the six poles with equal intensity suggests that the film is short of being single crystal and has two domains that differ by an in-plane rotation of 60°. The domain boundary areas will likely be the source of inhomogeneous carbon growth. The crystal orientation of the CoFe relative to the Al₂O₃ substrate is (111)CoFe//(0001)Al₂O₃ and [110]CoFe//[10 $\bar{1}$ 0]Al₂O₃. The FWHM of the (111) peak is calculated to be 0.2108° from rocking curve measurement, so this film is well oriented except for the in-plane rotation.

The CoFe fcc crystal structure should also be stable at low temperature, eliminating the problem of the complex multidomain formation of Co film due to the fcc-hcp transition in cooling. This should improve the homogeneity of graphene growth. However, iron does slightly increase the lattice parameter (table 1) and thus increases its lattice mismatch with graphene.

III. Ni film

The growth of 200 nm Ni film on c-Al₂O₃ at 500° C, 100 W and 3 mTorr of Ar was also investigated. The RHEED pattern taken after the growth showed nice long lines, indicating the surface is well ordered, which is reflected in its small FWHM value (0.3381°) of the (111) peak (table1). The results of the pole figure measurements showed that the film is face centered cubic. Similar to CoFe, the (200) pole figure of Ni (fig. 4b) showed six poles of similar intensity, indicating a two-domain structure with 60° in-plane rotation. The orientation relationship to Al₂O₃ is also (111)Ni//(0001)Al₂O₃ and [110]Ni//[10 $\bar{1}$ 0] Al₂O₃. Further improvement must be done to obtain single domain Ni films on c-Al₂O₃ or on (111) MgO for graphene growth. The M vs. H loop of the Ni film however showed a slanted slope (fig. 3c) with H_c of 101 Oe and M_r/M_s squareness value of 0.36. This is better than that of the cobalt film but still does not make it suitable as the bottom electrode for in-plane graphene MTJs.

IV. NiFe film

Similar to Co, the alloying of Ni with Fe can make it magnetically softer. Permalloy ($\text{Ni}_{80}\text{Fe}_{20}$) is known to have very small magnetocrystalline anisotropy with near zero magnetostriction. The M vs H measurements for the 200-300 nm Permalloy film on c- Al_2O_3 generally show an in-plane coercivity of 0.6 Oe with M_r/M_s squareness of 0.88, in comparison to 106 Oe and 0.36 squareness for the Ni film (table 1). This renders it suitable to be used as the bottom magnetic electrode for FM/Gr/FM spin valve structure. The lattice parameter of Py is larger than that of Ni, 3.54 Å vs 3.52 Å (table 1), and thus having larger lattice mismatch with graphene but with a lesser degree than that of CoFe. We therefore focus our efforts on the growth of Py film. Table 2 summarizes the results of our efforts to optimize the growth conditions for obtaining single crystalline Py film for graphene growth by CVD. The substrate temperature was varied between 300 to 600° C and a variety of single crystal substrates were used, including c-sapphire, (111)Si/(0001)AlN and (111) MgO.

First we compare the effects of temperature on a 300 nm Py film on c- Al_2O_3 . Films grown at 400° C have a large rms roughness of about 10 nm compared to the 1 to 2 nm for those grown at 500 and 600° C. The x-ray diffraction patterns (fig. 5) show that all films grown in the range of 400 to 600° C are highly (111) textured, but the lower temperature film has a detectable (200) peak of fcc Py (fig. 5a). These were absent for the films grown at higher temperature (fig. 5c). The results of the 500° C and 600° C samples are similar to each other, with the 500° C samples in general having higher (111) x-ray intensity and smaller rms roughness, making it a better growth temperature. The RHEED pattern of the 500° C film has nice long lines indicating good degree of surface order.

As that of the CoFe and Ni films, the Py on c- Al_2O_3 film is fcc. The pole figure of Py(200) also has six poles with the same intensity, meaning the formation of two domains with 60° in-plane rotation (fig. 4c). A phi rotation for the (111) poles at psi angle of 74.5° showed similar intensity peaks separated by 60° (fig. 6), again confirming the coexistence of two domains in equal amount. The orientation relationship of Py to Al_2O_3 is (111)Py// (0001) Al_2O_3 and [110]Py//[10 $\bar{1}$ 0] Al_2O_3 . The same orientation relationship was found for the film on the Si/AlN substrate.

In order to obtain single crystalline films, we experimented with first growing 50 nm film at 600° C and then decreased the growth temperature to 500° C to grow the remaining 250 nm film. This was to start the film with larger grain size but avoid the formation of deep grooves at 600° C for the rest of the film. We also grew film at 300° C for the first 50 nm and increased the temperature to 500° C for the rest of growth. Neither approach affected the formation of the two domain structure as evidenced from the pole figure measurements (not shown here). Sputtering with 5 W of rf bias applied to the substrate during deposition generated pits in the film. Annealing the films at 750° C for 2 hr in-situ after growth was found to develop a microstructure of large grains with deep grooves. The individual grains are smooth with rms roughness of 0.326 nm but the large grooves between them give the overall rms roughness the value of 3 nm for a 5 x 5 μm scan (table 2). The Py film grown on Si/AlN gave similar results as that on sapphire.

Cubic (111) MgO substrate was selected to avoid the formation of two different fcc stacking sequences on hexagonal substrate. From the phi measurement of the (111) Py poles, the usage of (111) MgO does effectively reduce the presence of second domain to less than 10%. The orientation relationship is (111)Py// (111)MgO and [110]Py//[110]MgO. Figure 6 compares the (200) pole figure for the 300 nm thick Py on c- Al_2O_3 , (111)Si/c-AlN, and (111) MgO. As can be seen, only three fold symmetry was observed for the film on MgO substrate. The FWHM of the (0006) Al_2O_3 , (0002) AlN and (111) MgO peaks are measured as 0.0098, 0.92, and 0.0167°, respectively. The large value of FWHM for the (0002) AlN peak may be a result of its silicon substrates having a 2 to 3 degree offset. The FWHM of the (111)

Py peak for these samples are all around or below 1 degree with the lowest value corresponding to the films on sapphire substrates (Table 2). This could be the effect of better substrate surface quality.

Figure 8 shows the AFM images of 300 nm Py film grown at 400, 500 and 600° C on c-sapphire and on Si/AlN. As can be seen, the 600° C samples have larger grain size and deeper grooves between the grain boundaries than the lower temperature samples. This indicates that the Py films at 600° C do not adhere well to the substrates, which seems to be worse for the film on Si/AlN. The rms roughness is larger (table 2) for the films on MgO. This is probably due to the ~16% lattice mismatch of Py (111) with MgO causing large line defects. The MgO substrates were found to be of poor quality, based on the results of x-ray diffraction and rocking curve measurements.

V. Graphene Growth

We experimented with CVD graphene growth using Co, Ni, CoFe and Py films with emphasis on Py on c-Al₂O₃. The samples were heated to 850- 900° C under Argon-Hydrogen atmosphere. They were exposed to Methane-Hydrogen atmosphere at 850-900° C for approximately 10 to 30 minutes before cooling to room temperature under Argon-Hydrogen atmosphere.

Figure 9 shows an AFM image of graphene on top of the Permalloy film. The AFM measurements show a surface roughness of 1.1 – 1.9 nm rms. This is partly attributable to wrinkles in the graphene film which emerge during cooling due to graphene's negative thermal expansion coefficient. Slip bands likely formed during cooling of the metal film are also visible in the AFM image and contribute slightly to the surface roughness. Its surface roughness, however, increased significantly more after CVD treatment. Single crystal graphene on Ni/(111)MgO was recently reported [16]. In this study the goal is to grow multilayer graphene on a suitable magnetic electrode and Ni did not have the right magnetic properties, whereas Py and CoFe did. Multilayer graphene CVD growth on Py was found to be of higher quality than that on CoFe films, although further improvements can be expected from changes to the film preparation prior to CVD.

The lattice parameter and magnetic properties of 300 nm thick Py film after graphene growth are listed in Table 3. Interestingly, the lattice parameter of Py was decreased after the graphene growth. If carbon were to remain in the matrix after the growth, then it should increase rather than decrease. This suggests that a small amount of iron diffused out of the matrix and formed iron carbide during the growth. This is not unexpected since iron does have a stronger reactivity with carbon than nickel. The increase of both the in-plane and out-of-plane coercivity of Py after the graphene growth indicates a change of microstructure and the formation of precipitates to pin the magnetic domain wall movement. From table 3, the (0002) carbon x-ray intensity was decreased for longer graphene growth time, suggesting the graphene formation is a dynamic process and the presence of hydrogen may etch away the already formed graphene layer [17].

Discussion

In order to obtain multilayer graphene forming a lattice matched structure with a Py or CoFe bottom magnetic electrode, growing a single crystalline or large grain bottom magnetic electrode is the first step. The surface preparation for the c-sapphire substrate appeared to be important for getting single crystalline films. High temperature annealing can remove the polishing scratches from the sapphire substrate and give an atomically smooth surface for the growth. Growing a seed layer of Al₂O₃ was reported to be effective in removing the OH layer from the surface of sapphire and reduced the domain formation in Ni films [12]. Annealing the magnetic film before graphene CVD is also crucial for obtaining large single crystalline grains. Some annealed at 500° C in Ar/H₂ for 3 hours [13], others

employed short period 1000° C annealing to obtain solid-phase epitaxial recrystallization (SPE)[7], while others reported covering Ni films with Ta foil (Ni film on c-sap is not stable above 900° C) when annealing at 800-900° C for one hour in UHV [18]. We intend to do more experiments to obtain single crystal FM films to grow large area epitaxial graphene.

The substitution of Fe to Co and Ni reduced the magnetocrystalline anisotropy and gives a large remanence with square magnetic switching suitable to be used as the bottom magnetic electrode for FM/Gr/FM spin valve structures. From the position of the fcc (111) peak, the lattice parameter of the CoFe is calculated as 3.551 Å. This corresponds to a value of 2.511 Å for the hexagonal in-plane lattice, comparing to 2.506 Å for Co and 2.45 Å for graphene. Similarly, the 20 at% substitution of iron to nickel increases the in-plane hexagonal lattice from 2.487 Å to 2.503 Å from Ni to NiFe. Thus the substitution of Fe to Co and Ni does increase the lattice mismatch with graphene but only by a small amount, so it may not affect the spin filtering effect. Theory prediction is still needed to know how alloying will change the electronic band structure of the FM and affect the graphene spin filtering.

As for the CVD growth, Fe does react more strongly than Co or Ni with carbon due to its having more vacancy in its d-orbitals [19]. This may affect the growth mode of graphene. According to reference #19, carbon solubility (atom%) at 1000° C for Fe, Co and Ni are 7.89, 3.41 and 2.03, respectively, thus the incorporation of iron should increase the dissolved carbon amount at high CVD growth temperature. Iron has a phase transition from fcc austenite to bcc ferrite below 912° C and the carbon solubility is then drastically decreased. The alloying of Co and Ni with Fe will likely change the growth dynamics of graphene by precipitation and so the CVD conditions such as the growth temperature and cooling rate would need to be optimized accordingly.

Summary

We investigated using 200-300 nm thick layers of Co, Ni, Co₉₀Fe₁₀ and Ni₈₀Fe₂₀ films as the substrate to grow 3 to 5 layers of graphene by chemical vapor deposition and as the bottom magnetic electrode for FM/Gr/FM based magnetic tunnel junctions. The magnetic films were grown by dc magnetron sputtering with different growth conditions onto c-Al₂O₃ substrates. The crystalline orientation, surface morphology/roughness and magnetic properties were measured using X-ray diffraction, AFM and VSM, respectively. Cobalt films grown at 500° C were found to be single crystal hcp and heteroepitaxial onto c-Al₂O₃. The epitaxial relationship is (0001)Co//(0001)Al₂O₃ and [11 $\bar{2}$ 0]Co//[10 $\bar{1}$ 0] Al₂O₃. The CoFe, Ni, and Py films on c-Al₂O₃ were found to be fcc and are (111) textured with domains having 60° in-plane rotation. The measured epitaxial relationship for these ferromagnetic films can be expressed as (111)_{FM}/(0001)Al₂O₃ and [110]_{FM}/[10 $\bar{1}$ 0]Al₂O₃.

The magnetic properties showed small remanence and large coercivities for the Co and Ni films and large remanence and small coercivities for the CoFe and Py films, making the latter two films more suitable to be used as the bottom electrode for CPP device with in-plane magnetization. The c-axis oriented Co film however is a good candidate to be patterned into small devices that need a perpendicular magnetic anisotropy (PMA) material. Since the domain boundary area is a major source of inhomogeneous graphene growth, different sputtering conditions and substrates were experimented in order to obtain single crystalline Py film for growing large area of epitaxial graphene having uniform thickness. Py films grown on (111)Si/(0001)AlN were found to behave the same as those on sapphire, having in-plane rotational disorder. The Py films on (111) MgO were found to be mostly single domain.

The lattice parameter of Py is larger than that of Ni thus it increases the lattice mismatch with graphene, which may affect its spin filtering effect. The higher solubility of carbon in Fe at the CVD

temperature may change the precipitation and growth dynamics of graphene. In addition, Fe reacts stronger with carbon to form iron carbide which can affect the growth mode of graphene. We found that the lattice parameter of Py was reduced after graphene growth, suggesting the diffusion of iron into grain boundary area to form iron carbide. This may be the cause for the higher in-plane and out-of-plane coercivity of the Py films after graphene growth.

Acknowledgements

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List of Tables

Table 1. The lattice parameter and magnetic properties of the c-Al₂O₃/FM films.

film	(hkl)	2 theta(°)	d(Å)	a _{fcc} (Å)	a _{hcp} (Å)	FWHM (°)	H _C (in-plane, Oe)	s (M _r /M _s)
Co	0002	44.54	2.033	3.544*	2.506*	0.3277	139	0.21
CoFe	111	44.141	2.05	3.551	2.511	0.2108	4.6	0.97
Ni	111	44.567	2.031	3.518	2.487	0.3381	101	0.36
NiFe	111	44.282	2.044	3.540	2.503	0.1973	0.6	0.88

*ref. [20].

Table 2. The results of Permalloy films grown at different conditions and on different substrates

substrate	T(°C)	treatment	t (nm)	rms (nm)	FWHM (°)	H _C (in-plane, Oe)	H _C (out-of-plane, Oe)
c-Al ₂ O ₃	400		~320	10.7	0.51	4.95	135.2
(111)Si/c-AlN	400		~320	7.8	1.31	4.3	7.7
c-Al ₂ O ₃	500		~300	4	0.22	1	
(111)Si/c-AlN	500		~300	0.9	0.57		
(111)MgO	500		~300	1.1	1.07	1.2	
c-Al ₂ O ₃	500		326	1.5		1.17	58.6
(111)Si/c-AlN	500		326	1.3		1.26	34.4
c-Al ₂ O ₃	600		318	1.7	0.25	1.08	28
(111)Si/c-AlN	600		318	1.8	0.38	1.14	68
c-Al ₂ O ₃	650/500		297	1.4	0.2	0.94	33
(111)Si/c-AlN	650/500		297	1.2	0.42	1.02	79
(111)MgO	650/500		297	2	1.23	1.37	29
quartz	650/500		297			5.2	83
c-Al ₂ O ₃	300/600		194	1.9	0.29	1.09	17.93
(111)Si/c-AlN	300/600		194	1.6	0.71	1.06	143
c-Al ₂ O ₃	500	750 °C, 2hr	323	3	0.23	1.11	21
(111)Si/c-AlN	500	750 °C, 2hr	323	2.4	0.33	1.25	27.5

Table 3. The effect of graphene growth on lattice parameter and magnetic properties of 300 nm Py on c-Al₂O₃.

t of Gr growth (min)	Py (222) (°)	d of Py (222) (Å)	a _{fcc} (Å)	a _{hcp} (Å)	I _{Gr} (0002) (count/s)	H _c (in-plane, Oe)	H _c (out-of-plane, Oe)
0, pristine	97.79	1.0223	3.5413	2.5041		0.98	13.5
10	97.81	1.0221	3.5408	2.5037	862	3.02	47.2
20	97.83	1.0220	3.5402	2.5033	1461	2.97	88.2
35	97.87	1.0217	3.5392	2.5026	308	2.74	94.9

List of Figures

Figure 1. The 2 theta-omega scans of c-Al₂O₃/300nm Co (red curve) and Si/SiO₂/7nm Ta/10nm Ru/300nm Co (blue curve). The x-ray intensity was plotted in square root scale.

Figure 2. (a) (10 $\bar{1}$ 3) pole figure of 300 nm Co film on Si/SiO₂/7nm Ta/10nm Ru, The red ring at psi of 30° shows that this film has a strong out of plane texture with random in plane orientation; (b) (10 $\bar{1}$ 3) pole figure of 300 nm Co film on c-Al₂O₃, the six inner poles are that of (10 $\bar{1}$ 3) poles of Co while the six outer poles are that of Al₂O₃; and (c) (10 $\bar{1}$ 1) pole of 300 nm Co on c-Al₂O₃.

Figure 3. M vs H curves of the ferromagnetic films on c-Al₂O₃ substrate a) 300 nm Co b) 200 nm Ni c) 200 nm Co₉₀Fe₁₀ and d) 200 nm Ni₈₀Fe₂₀.

Figure 4. (200) pole figure of 200 nm film on c-Al₂O₃ substrate of a) Co₉₀Fe₁₀ b) Ni and c) Ni₈₀Fe₂₀.

Figure 5. The x-ray diffraction pattern of 300 nm Py film on c-Al₂O₃ grown at (a) 400 (b) 500 and (c) 600° C. The x-ray intensity was plotted in arbitrary unit and in log scale.

Figure 6. Phi scan of (111) Py on c-Al₂O₃ at psi angle of 70.5°. The five peaks with identical intensities 60° from each other shows that the two domains occur in equal amounts.

Figure 7. (200) pole figures of 300 nm Py on (a) c-Al₂O₃ (b) (111)Si/(0001)AlN and (c) (111) MgO.

Figure 8. AFM images of 300nm Py on c-Al₂O₃ grown at (a) 400 (b) 500 and (c) 600° C and on (111)Si/(0001)AlN grown at (d) 400 (e) 500 and (f) 600° C. The images are for 1 x 1 μm scan size.

Figure 9. AFM image of c-Al₂O₃/300nm Py/Graphene. The tall linear features are attributed to wrinkles in the overlaid multi-layer graphene film. Surface roughness is 1.1 – 1.9 nm rms.

Figure

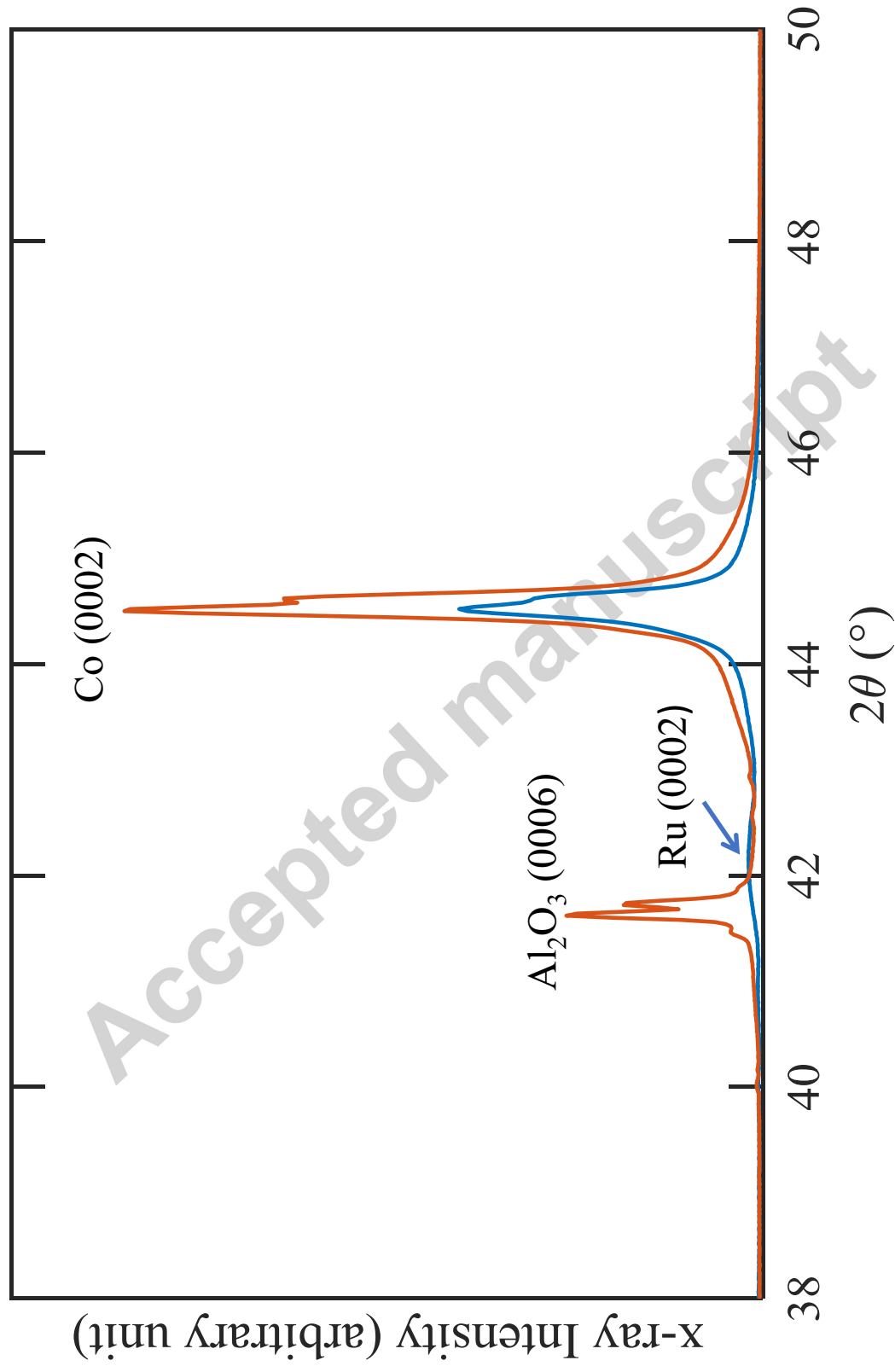


Fig. 1

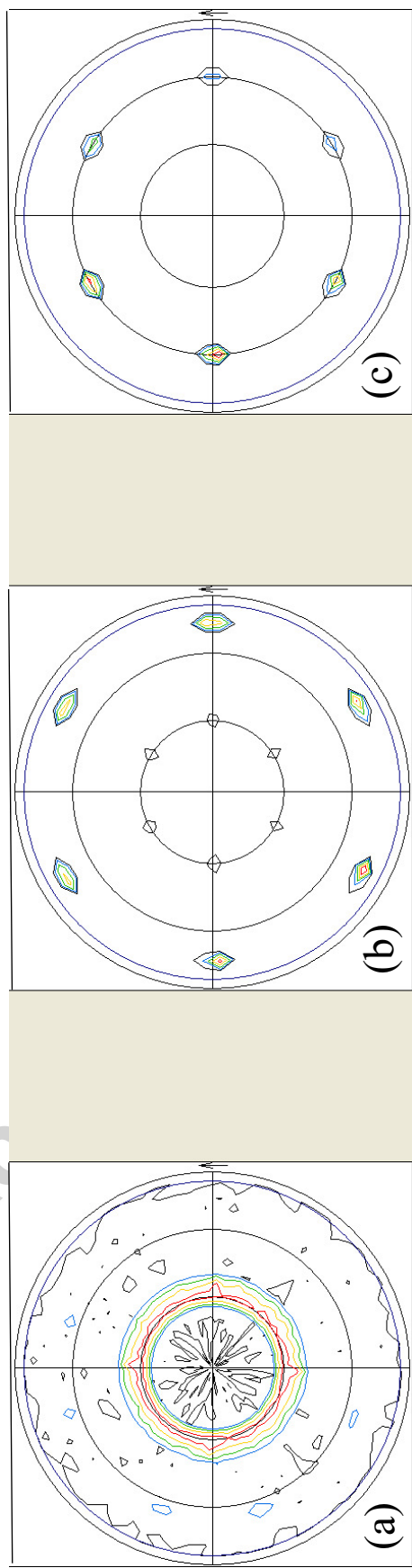


Fig. 2

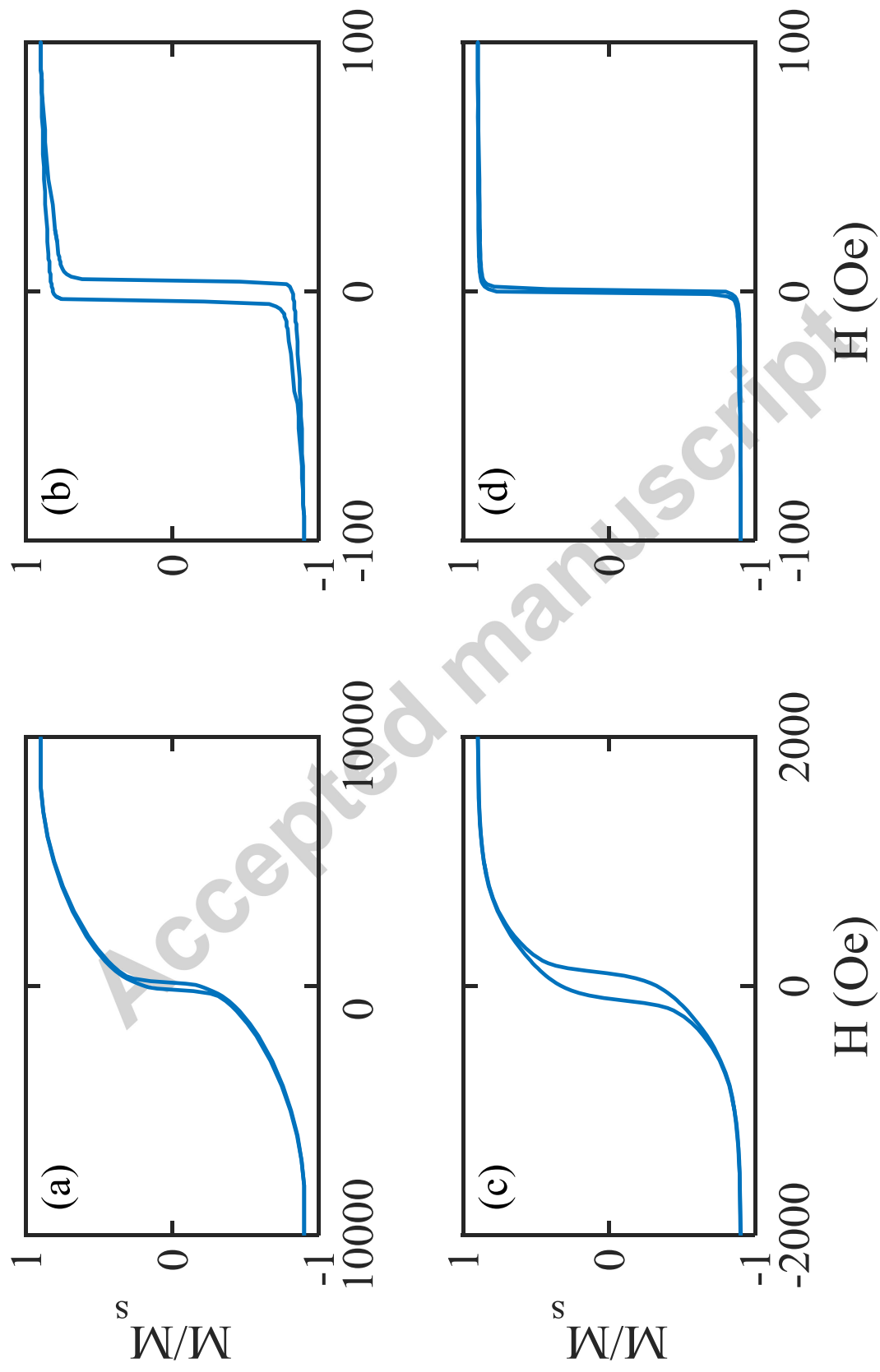


Fig. 3

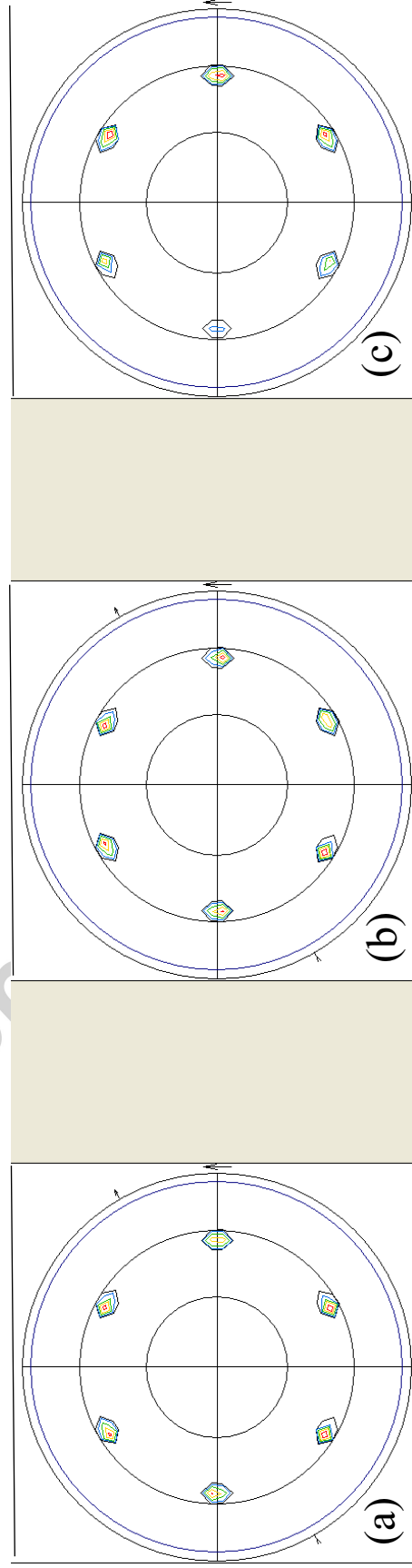


Fig. 4

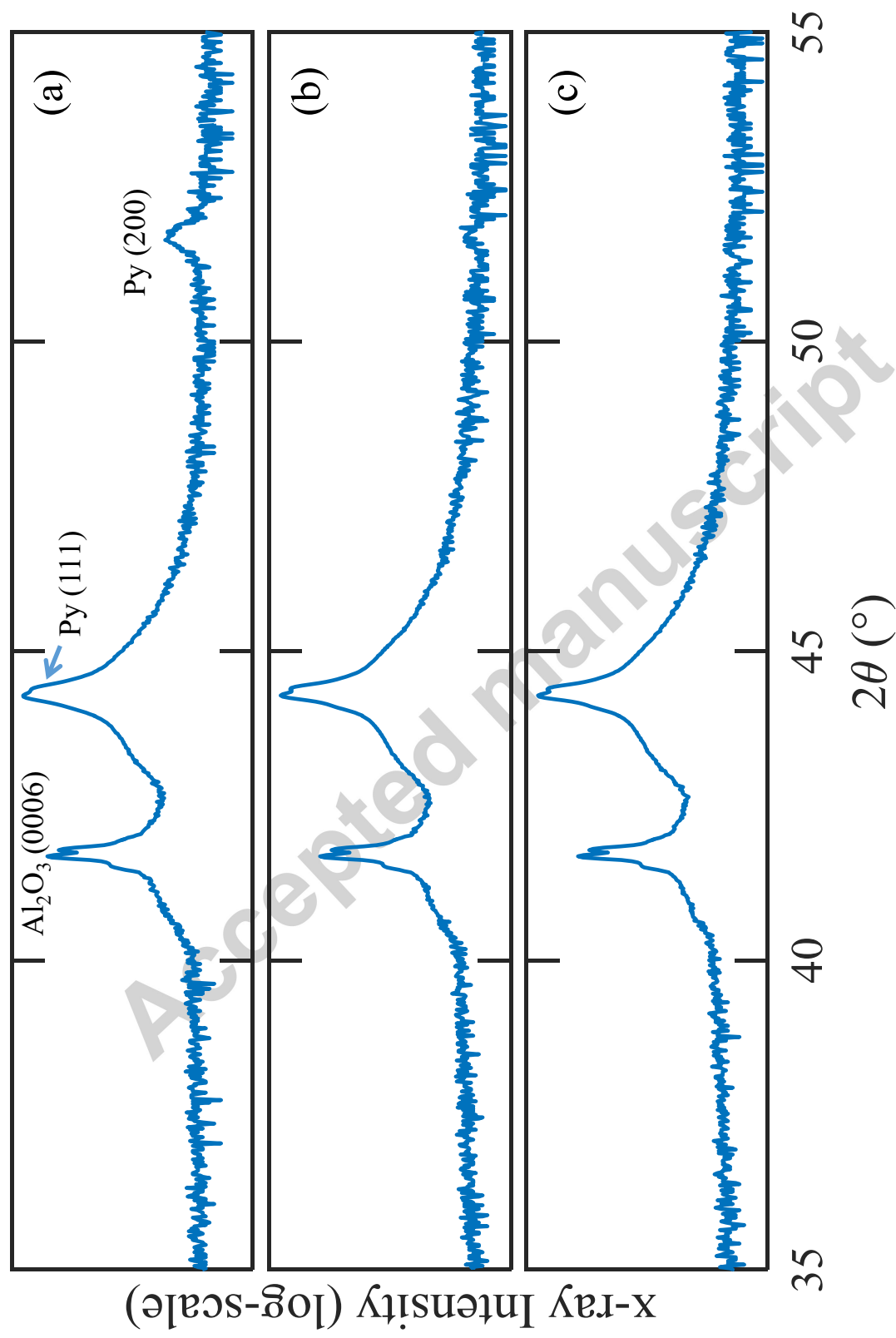


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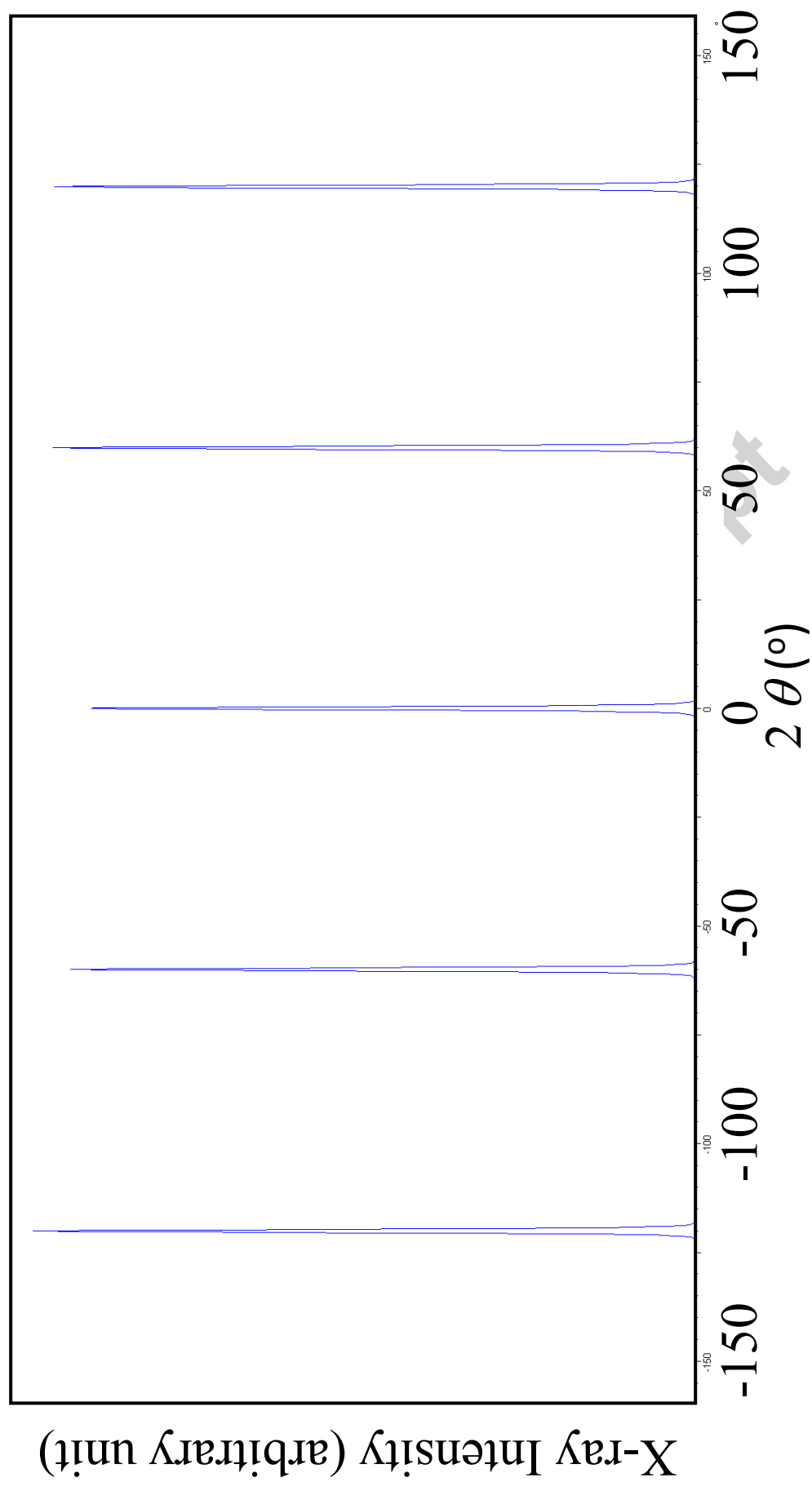


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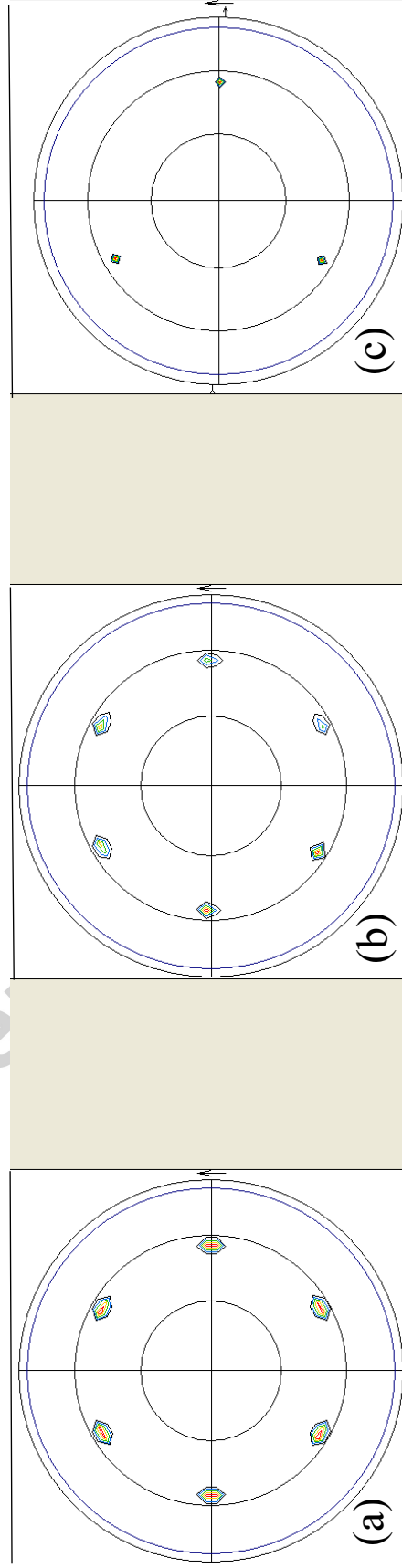


Fig. 7

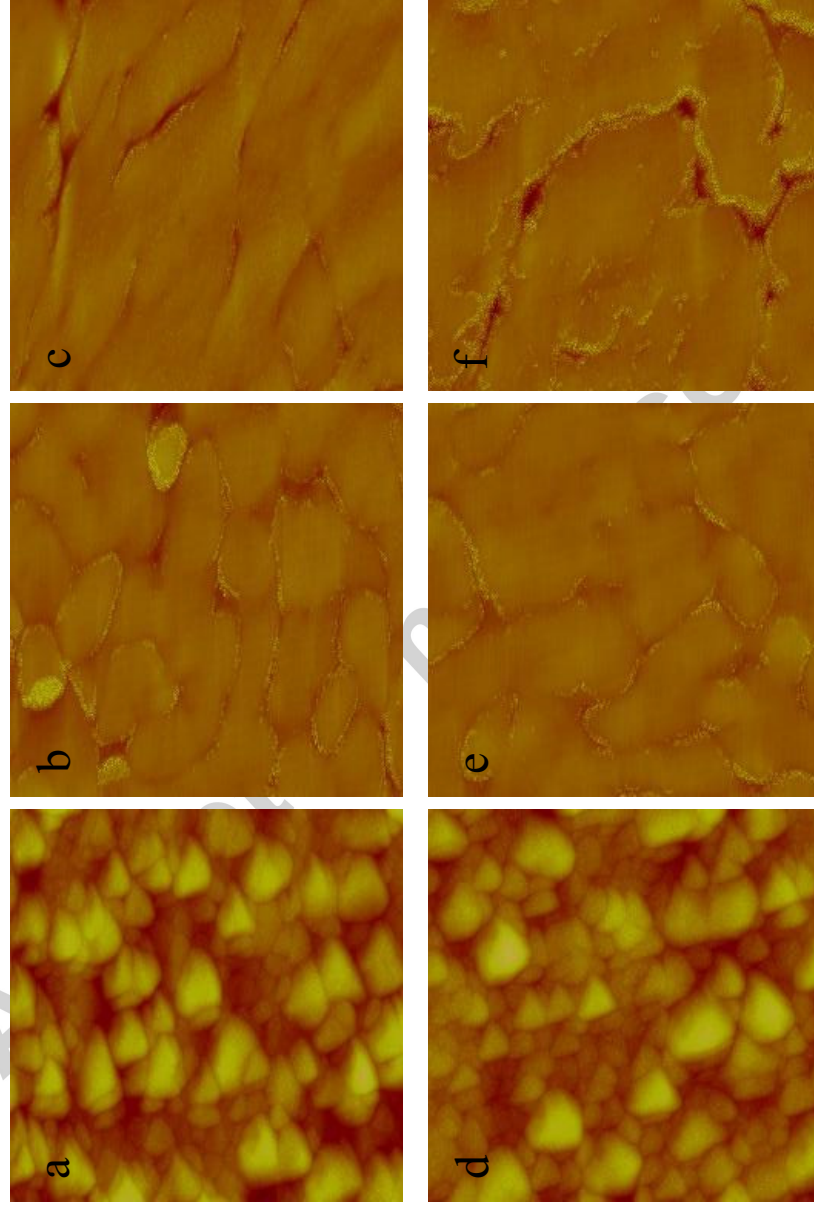


Fig. 8

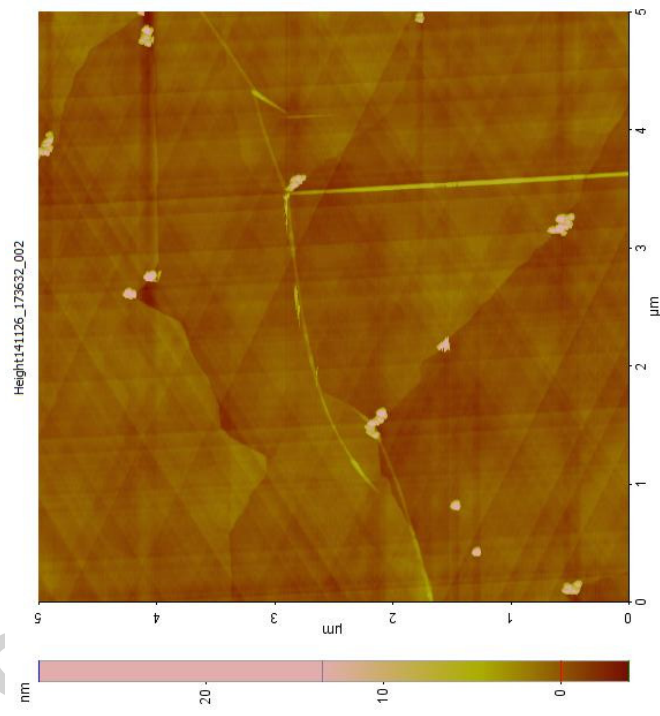


Fig. 9